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Topology optimized lattice core effects on the flexural behavior of Ti-6Al-4V sandwich structures produced by electron beam powder bed fusion (EB-PBF)

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Abstract

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Metal sandwich structures are utilized in aerospace applications that require high specific flexural strength and temperature resistance. Conventional fabrication methods, which involve joining operations (e.g. brazing, welding, bonding) to attach face sheets to a core structure, limit the geometric complexity that can be produced economically with sufficient quality. New additive manufacturing (AM) technologies, such as electron beam powder bed fusion (EB-PBF), present the ability produce a monolithic sandwich structure from Ti-6Al-4V with novel topology optimized cores. For this investigation, five lattice core geometries were selected, three triply periodic minimal surface (TPMS) types: Gyroid, Diamond, and Primitive, and two strut types: Octet and Kelvin. To demonstrate the unique capability of AM to optimize structures, two methods for functionally grading the core densities were applied based on a 4-point bend

simulation. One grading method used a compliance minimization strategy (density-based), and the other a von Mises stress minimization strategy. After using waste powder from another additive process to produce 15 novel sandwich structures, the flexural properties were determined by ASTM C393 4-point bend experiments. This research revealed that the continuous grading of the core densities successfully delayed flexure yielding without increasing weight, and the TPMS core geometries exhibited a higher resistance to core buckling. The stress minimization strategy provided superior performance, while the Diamond core geometry attained the highest flexure yield stress. This work serves as a valuable contribution in the design for additive manufacturing of metal sandwich products, and highlights the sustainability of the EB-PBF process.

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NOMENCLATURE

3D	Three Dimensional
AM	Additive Manufacturing
ASTM	American Society for Testing and Materials
CAD	Computer Aided Design
CMOS	Complementary Metal Oxide Semiconductor
DED	Direct Energy Deposition
DfAM	Design for Additive Manufacturing
DfMA	Design for Manufacturing and Assembly
DMLS	Direct Metal Laser Sintering
EB-PBF	Electron Beam Powder Bed Fusion
FEA	Finite Element Analysis
FEM	Finite Element Model
FFF	Fused Filament Fabrication
HIP	Hot Isostatic Pressing
L-PBF	Laser Powder Bed Fusion
PBF	Powder Bed Fusion
PRS	Powder Recovery System
SLM	Selective Laser Melting
SLS	Selective Laser Sintering
STL	stereolithography
UTS	Ultimate Tensile Strength
YTS	Yield Tensile Strength

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DEDICATION

I would like to dedicate this work to my loving wife, Raquel, my son, Wyatt, and daughters, Jasmine and Aurora, because without their support, flexibility, and grace, this contribution to science would never have been possible. They are the ones that give this work, and the time invested, real value and meaning.

Chapter 1. INTRODUCTION

In industry today, to manufacture a traditional complex assembly of many parts, the engineering requires individual component drawings, as well as assembly drawings, and planning documents for how to assemble them. Each part in that assembly, fabricated by conventional means, will typically require additional detail level drawings and planning documents. Those detail parts will then require tools and fixtures for fabrication, and each tool requires their own drawings, materials, and planning documents. Added together, the total cost of materials, labor, tooling, engineering drawing releases, and lead time for the assembly, totals to significant amounts. However, large recurring and non-recurring savings could be realized if that assembly could instead be a single component, additively manufactured on a single machine, in only a few days, and accomplished right alongside multiples of other unique complex parts. This highly desirable future state of additive manufacturing (AM) is rapidly becoming a reality in many industries, and is a reason for the increasing pull for more research and development.

Most of the expansion of the field of AM research has taken place over the past decade as shown by a 2020 bibliometric analysis conducted by Patil and Soni [1]. They searched the Scopus database for “additive manufacturing,” or “3D printing,” or “rapid prototyping,” and over a date range from 1970 to 2019 there were 49,508 publications found, as seen in Figure 1.1, which were then broken down into academic fields, countries, journals, and authors. More than 50% of the publications were issued in the field of engineering and materials science, with the top three countries being the US (14,548), China (6,855), and Germany (4,270). The top three journals were Rapid Prototyping, Intl Journal of AM technology, and Additive Manufacturing, where the top US author being D. W. Rosen from Georgia Tech, ranked 3rd globally.

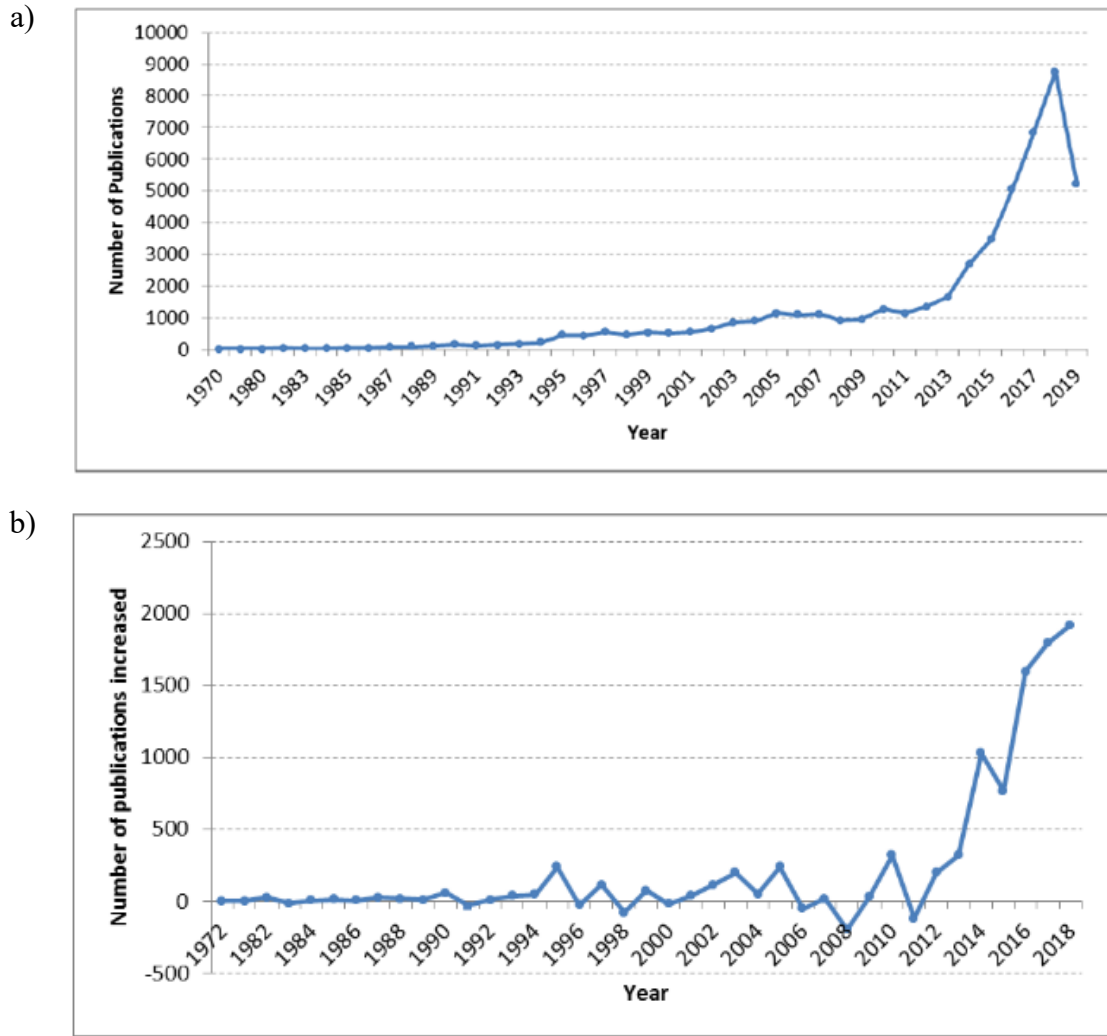


Figure 1.1. a) Number of publications per year found in the Scopus database when searched for “additive manufacturing,” or “3D printing,” or “rapid prototyping,” and b) the number of publications per year. [1].

Industries such as medical, aerospace, and automotive, that depend on lightweight, complex products are investing heavily in AM since weight is a key metric to product performance and operating costs. That is driving the direction of research to understand the various additive processes and parameters, in order to reach a stable and repeatable manufacturing state. While there are many benefits with AM being a direct digital manufacturing method, security has become

a concern as the process can also be a potential target for cyber-physical attack. An example provided by Graves et. al., [2] showed how a feedstock delivery subsystem could be manipulated to degrade the mechanical properties of parts. This study illustrated the sensitivity of the process and the difficulty in detecting an attack that reduces the quality of the produced components.

A review by Dejene and Lemu [3] pointed out that as AM technologies have gained increased attention with their manufacturing capabilities, the focus of discussion has turned more to the reliability and quality of the parts produced. They stated that there is a consensus amongst experts and stakeholders from the medical, aerospace, and automotive industries, that metal AM components still do not meet many of their strict quality requirements. Particularly in regards to overall equipment effectiveness (OEE), requiring $> 70\%$, and scrap rates < 1000 ppm, indicating that repeatability and reproducibility limitations are among the greatest barriers. This article acknowledged that while there is a large quantity of AM research investigating the technology capability, repeatability and reproducibility focused research has been lacking. However, those two components, along with dimensional accuracy, and precision, are key product quality and reliability traits.

Two surveys were conducted regarding the necessity of quality strategies in metal AM technology, and the study results were recently presented in 2025 by Sithole et. al [4]. Those surveyed were experts in metal AM technologies from academic and industrial backgrounds, and attendees of Germany's largest annual AM event, Formnext. Key challenges identified were material inconsistencies, process variability, and surface finish issues. These highlighted the need for robust quality strategies that focused on feedstock quality, optimization of parameters, and defect detection. As shown in Figure 1.2, Design for AM was identified as one of the biggest components of an effective quality strategy, with ISO 9001 in a close second, which ensures

production processes are well-defined and controlled with consistency and traceability. Quality frameworks with integrated quality management, need to be designed that will foster reliable AM processes that are scalable, and will emphasize defect prevention more so than defect detection.

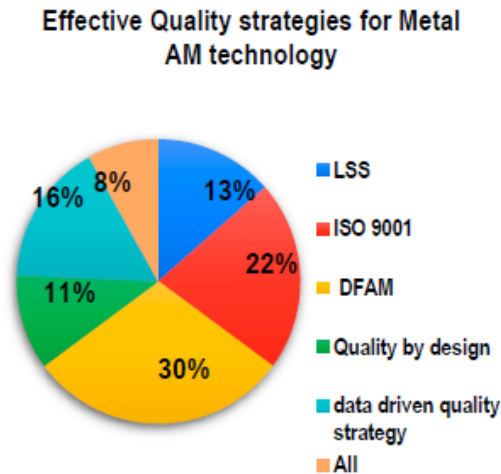


Figure 1.2. Most effective metal AM Quality strategies according to Formnext event attendees [4].

A 2022 review of qualification and certification for metal AM was published by Chen et. al. [5]. While in many cases the terms “qualification” and “certification” are often used interchangeably, this article clarified the terms such that qualification was defined to be the evaluation of something (e.g. the design, material, product, technique, machine, or service) during development or testing, and certification was meant to satisfy an authority at the conclusion of development. The article covered the progress made in the ongoing standardization efforts by 16 different organizations, particularly ISO and ASTM whom have led the way with 28 published international standards and 60 in-development. These standards cover such things as terminology, general principles, materials and processes, design, test methods, machine qualification, and data (e.g. AM file format), though the topic of qualification in general is significantly lacking in the quantity of published standards. Critical challenges for the qualification and certification of metal

AM are in the aspects of materials, traceability, repeatability, design, post-processing, and in-situ monitoring, with high cost and prolonged lead times being cited as some of the more significant impediments. The lack of in-situ monitoring for real-time process control is also a substantial gap to obtaining process reliability and repeatability.

In the review [5], a qualification and certification framework was proposed which required more expertise in metal AM than the more universal quality management systems (QMS), such as ISO 9001 and others. This framework for metal AM consisted of specific standards, rules, and regulations that addressed the technology, design, materials, printing, post-processes, inspection, and testing, in order to provide a consistent approach to guarantee quality products and services. Since the qualification and certification methodologies of traditional manufacturing processes have required high cost, long lead times, and substantial testing, the outlook for metal AM should focus more on digital methodologies as part of an industrial digital transformation. In this way, the minimization of defects and increased product performance can be accomplished with minimal human interaction, and relieve many of the final burdens with efficiency and transparency.

The topic of titanium additive manufacturing methods, properties, challenges, and prospects was covered by a 2022 review [6]. Titanium alloys are desired, particularly in industries such as biomedical, aerospace, defense, and automotive, because of their specific strength (Figure 1.3), hardness, wear properties, biocompatibility, corrosion resistance, and low modulus of elasticity; however, those properties also make the processing of titanium by conventional means more challenging. The most common AM fabrication methods were reported to be electron beam and laser powder bed fusion (EB-PBF and L-PBF), laser based directed energy deposition (DED) with a powder feedstock, and binder jetting. In addition to the most common titanium alloy, Ti-6Al-

4V, the article discussed investigations performed with TiAl, which is gaining interest for its improved oxidation resistance that enables higher temperature applications.

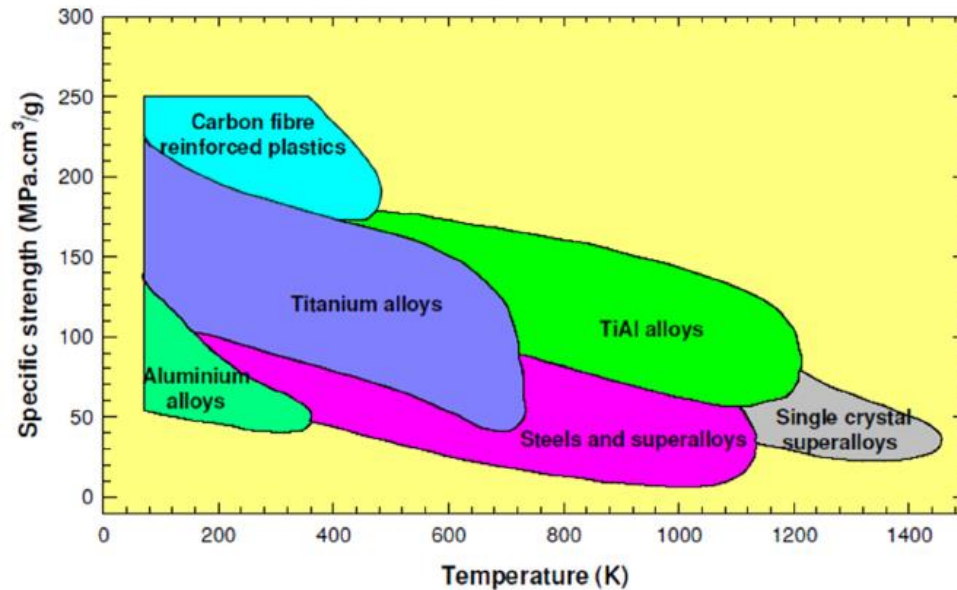


Figure 1.3. Specific strength comparison of titanium alloys with other structural materials as a function of temperature [7].

As industrial acceptance continues to grow with more research and understanding, how PBF technologies will integrate with well-established design, manufacturing, and quality processes remains a challenge. In a review of PBF for aerospace applications by Katz-Demyanetz et. al. [8], the authors identified common development trends. One of these trends was expanding the range of materials such as metallic alloys, metal or ceramic matrix composites, and functionally graded materials. Another trend was to increase the rate of various PBF techniques in aerospace application trials, or with a focus on critical aerospace components. Developing new methods of quality control, post-processing, and standardization for AM components is a serious undertaking that involves new approaches to equipment, work-force training, logistics, and resource management. Thus, to realize the full potential of PBF in aerospace, future directions should

include: the production of lightweight defect-free parts, higher performing alloys with advanced microstructure, and composite materials optimized for a specific function.

A review of AM processes in medical and dental applications was published by Salmi in 2021 [9]. It identified the numerous reviews that have already been written on the use of AM in the medical industry, yet noted that only in limited cases were the particular AM processes described. Powder bed fusion (PBF), material extrusion (FFF) and VAT photopolymerization or stereolithography (SLA), are well established for all categories of medical and dental applications, while sheet lamination (SL) and directed energy deposition (DED) applications are rare. Binder jetting and material jetting are also well established, but their applications are largely limited to medical models. The future direction of medical AM research was stated as process dependent and focused on those processes that are not as widely used. For instance, more investigations would be useful in the repair of medical devices and tools using DED, while multi-metal sheet laminated parts may also have promise for medical instruments, tools, and devices. Multi-material composite parts produced by material extrusion can serve as splits, prostheses, supportive guides, and other medical aids. Investigations into biomanufacturing using material jetting need to be expanded, and binder jetting of metal parts needs further development and understanding.

Another 2021 publication reviewed the additive manufacturing development of Ti-6Al-4V implants for the biomedical industry via EB-PBF [10]. From the 360 sources cited, the authors Tamayo et. al., covered the various PBF techniques before focusing on electron beam and the assortment of alloys that have been researched for medical applications. Many of these studies produced very small porous structures and characterized the various defects that reduce the mechanical properties. The primary material discussed was on how titanium, and how the ($\alpha + \beta$) dual structure of Ti-6Al-4V provides good fatigue strength and high hardness (i.e. wear resistance).

The alloy also provided a decent balance between physiochemical properties and biological functionality. Where high surface roughness and porosity are typically undesirable and detrimental to mechanical properties, those characteristics are favored by living tissue that needs oxygen, nutrients, physiological fluids, surface bioactivity, osseointegration, vascularization, and cell proliferation and adhesion processes. The review covered a variety of post processing methods such as hot isostatic pressing, machining, and anodizing, which can assist with optimizing performance within the body. Several case studies were reported that included a knee implant and jaw implant; however, no analytical models or numerical methods were discussed.

Quality issues and mitigation strategies in metal PBF were reviewed by Ravalji and Ravel [11], which further stressed the need for in-situ process monitoring and feedback controls to avoid the waste of time and money producing defective parts. Only the most common materials such as AlSi10Mg, Ti-6Al-4V, 316L Stainless Steel, and Inconel 718 have been widely reported on for parameter optimization and quality issues, yet these may not be the optimum alloys for the PBF process. More experimental investigations are needed on the effects of powder recycling and feedstock management on fabrication quality, plus the multi-physics finite element tools need higher fidelity and larger scales to adequately predict the risk of defect generation. There is the potential for machine learning algorithms to strengthen quality control through the processing of large quantities of data generated by in-situ build monitoring of surface roughness, dimensional accuracy, part density, and residual stress.

It is often stated that PBF processes are expensive and slow. That a failure occurring near completion of a large build, or is discovered after completion, incurs a high cost. Therefore, many in the AM industry feel in-situ process monitoring is critical to quality assurance with the detection of defects as they occur and under what conditions. Di Cataldo et. al. published a recent review

pertaining to quality inspection and control of metal PBF, which explained the challenges with monitoring all the different length scales [12]. As shown in Figure 1.4, the authors identified melt pool sensing, scan path sensing, slice level sensing, powder bed sensing, and full build sensing as the different scale levels needed for full monitoring.

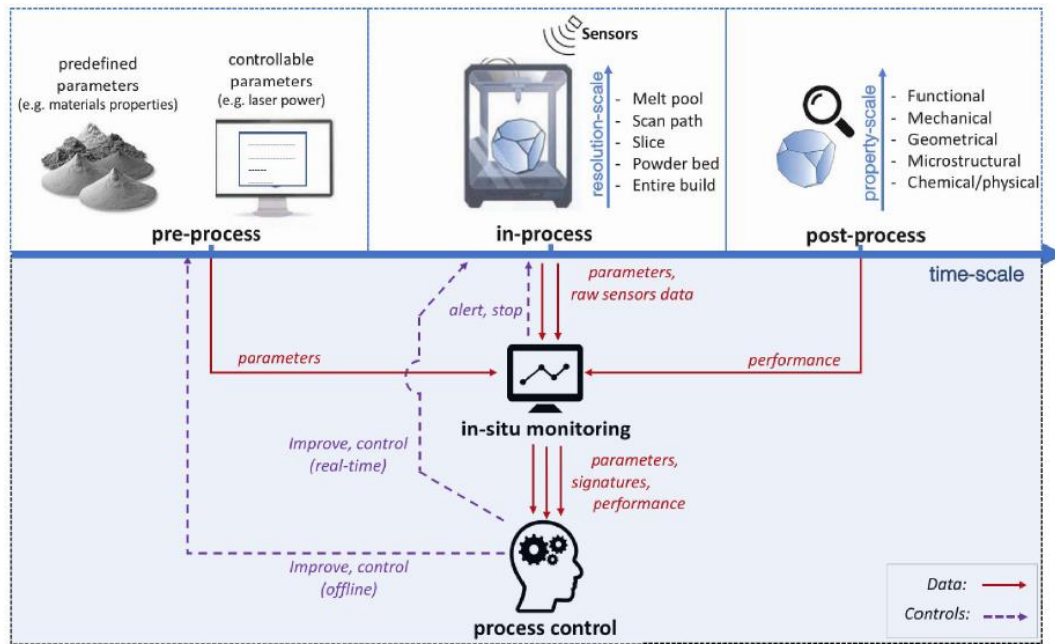


Figure 1.4. Quality assurance in PBF AM conceptual flow according to Di Cataldo et. al. [12].

Each level though required different equipment, and thus outputs various data formats. An example was provided for a laser melt pool monitoring system, at a typical scan speed, the melt pool will persist for about 0.001 of a second, with heating and cooling rates of about 10^6 K/s, a system capable of capturing the melt pool dynamics would require at least a 10 kHz collection rate. Now scale that over a build that takes 3 days to complete and the data management quickly becomes overwhelming. Even at the slice level where collection rates will be more manageable, a large 100 x 100 mm field of view, will require a resolution of $\mu\text{m}/\text{pixel}$. In-situ monitoring was noted to be even more difficult for EB-PBF with very limited access to the vacuum chamber and

high build temperatures that would destroy most sensors. The focus of the article then turned to machine learning applications to process the data through neural networks to help identify potential build issues. The current challenge there is on establishing the training sets with which the algorithms can determine good, from potential problem. While the future appears optimistic for this space to develop into practical applications of quality assurance, it is still a fairly distant Horizon 3 type technology, and metal PBF will need to rely on existing post-process inspection methods to determine part quality.

With metal PBF technology, it is now possible to manufacture designs that were previously impossible. In the past, engineers using traditional 3D computer aided design (CAD) software tools could create designs that could not be manufactured in a cost-effective way, but now the complexity capability of PBF is such that it can manufacture structures that cannot be explicitly designed. Architected cellular lattice structures for example are a significant challenge for traditional CAD software packages to model, which is where new implicit modeling methods like those employed by nTopology, have demonstrated capability. Additionally, the field of topology optimization has been expanding, now that the organic lightweight shapes it generates can be more easily produced with AM machines.

As stated by Seepersad [13] when discussing the challenges and opportunities in design for AM (DfAM), it was stated that in order to unlock the full potential of AM, the way engineers approach design has to change. That conventional fabrication techniques, such as injection molding, casting, and machining, have created conceptual barriers that must be broken down. At this point in time, engineers have almost exclusively had to follow all of the design for manufacturing and assembly (DfMA) guidelines and restrictions that accompany conventionally fabricated parts. Designers experience a dominant tendency to adhere to solutions they have

encountered previously, which makes it difficult when presented with the blank volume of an AM machine to conceive of marketable designs that could not be made another way. Bioinspired examples from nature and topology optimizations, coupled with education initiatives, are needed in order to change perspectives and shift mindsets. For this reason, it is important to introduce new generations of engineers to the unique tools and skills necessary for successfully DfAM implementation.

Topology optimization can help in the design of lightweight structures because it leverages a material's capability in the most effective way. It offers a mathematical framework to determine the most efficient layout, connectivity, and shape for prescribed objectives, constraints, material behaviors, and operational conditions. The components that are most central to the optimization formulations are: the definition of objective or cost functions, constraints dictated by physics or other requirements that bound the design variables, analysis tools for the problem of interest, the algorithm solution strategy, and a method for describing an evolving topology. The combination of additive manufacturing and topology optimization has created opportunities for changing the overall manufacturing and business models, particularly in the fashion industry [14]. There can be a more sustainable production model based on the ease of design customization and AM that uses recycled waste or biodegradable materials. Although, one of the primary limitations is also the small number of materials currently available, many of which lack the high flexibility and durability to compete with materials used in traditional manufacturing processes.

According to a recent review article published in 2023 regarding topology optimization for metal additive manufacturing, Ibhade et. al. [15], discussed the current trends, challenges, and future outlook. While topology optimization was originally mostly theoretical due to manufacturing limitations that prevented actualizing such designs, there has been a lot of effort to

incorporate topology optimization into the design workflow thanks to advancements in AM. Numerous methods of topology optimization have been developed which are categorized into density-based methods (e.g. Solid Isotropic Material with Penalization, and Rational Approximation of Material Properties), evolutionary methods (e.g. Evolutionary and Bi-Evolutionary Structural Optimization), boundary variation methods (e.g. level set, and phase field), or non-gradient-based methods (e.g. Genetic Algorithm, Particle Swarm Optimization, Simulated Annealing, and Artificial Intelligence Assisted). With the consideration of additive manufacturing, the application of topology optimization has expanded beyond just the desired part geometry. It can also assist with optimizing build orientation, the design of support structures, and reducing thermal accumulation and residual stress, in order to prevent undesirable distortion while limiting material usage and post processing. Most of the progress towards application of topology optimization has been made in the aerospace, medical, and automotive industries, though the most significant challenges remain in the areas of standardization and certification.

An extensive review by du Plessis et. al., published in 2021, focused on the properties and applications of metal additively manufactured architected cellular lattice structures [16]. Lattices create an inherently very high surface area to volume ratio because they are porous with highly interconnected pore spaces. The authors defined the terms “lattice structure,” “cellular structure,” and “architected cellular materials” as interchangeable, all referring to purposefully designed additively manufactured porous structures. These terms are not only used to describe strut-based topologies, but also include Triply Periodic Minimal Surfaces (TPMS), and “auxetic” (negative Poisson’s ratio) architectures, examples of which are displayed in Figure 1.5. They also point out that for lightweighting, lattice structures offer high stiffness-to-weight ratios, but that their stiffness is relatively low compared to a parent material solid.

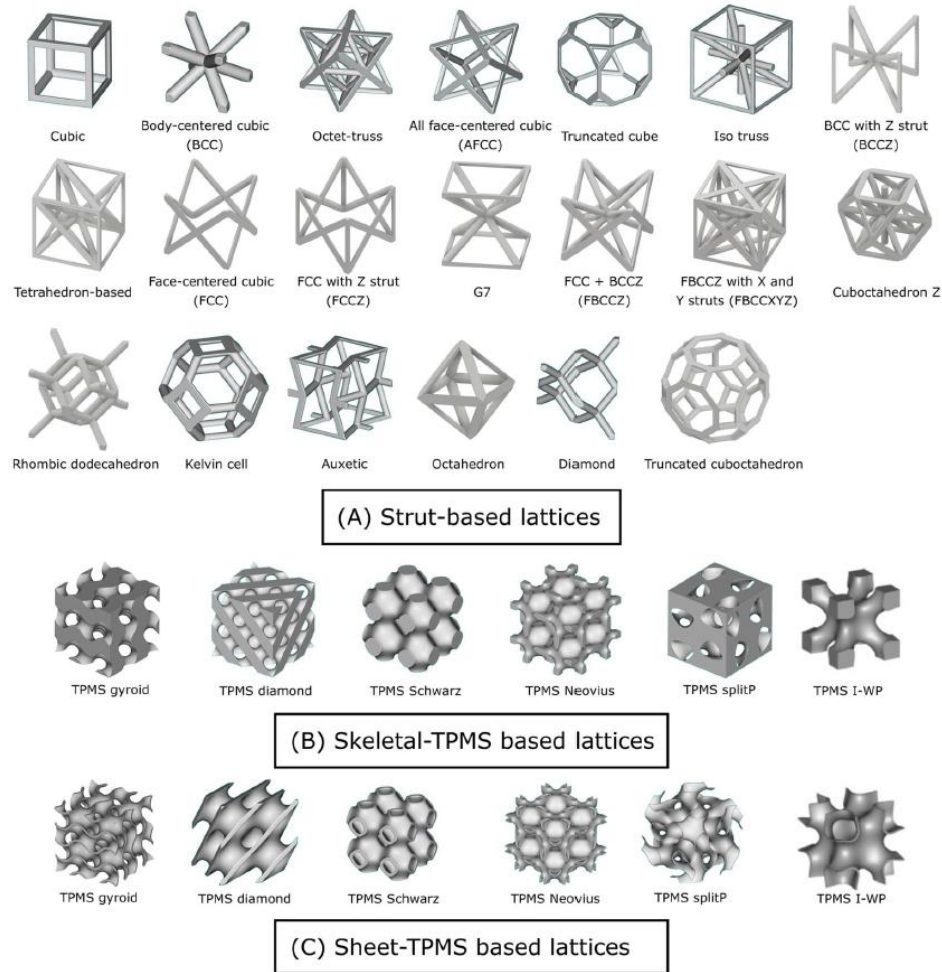


Figure 1.5. Lattice architectures grouped by type, showing various examples of strut-based unit cells, and triply periodic minimal surface (TPMS) skeletal and (C) sheet unit cells [17].

Research studies have therefore combined lattice cores with skins of various materials and thicknesses to form light weight sandwich panels [16]. However, one aspect the authors clearly identified as lacking in available literature was in the analysis of the flexural behavior of lattice structures. Most experiments are simply conducted in compression, while flexural experiments with 3-point or 4-point bend loading conditions have practical challenges in applying concentrated loads to a lattice material that lead to indentation. Therefore, these tests have been largely limited to sandwich structures where fully dense face sheets, that are able to withstand the load application

and reaction stress concentrations, serve to distribute the loads into the lattice core. Some of the more common TPMS lattice geometries mentioned were the Gyroid, Diamond, and Schwarz Primitive.

This review also identified the challenges associated with DfAM strategies [16]. AM machines are capable of producing complex geometries such as lattices, but defining the geometries in traditional explicit CAD design software is not trivial. The explicit definition provides direct control of node locations and surfaces in 3D space, but for lattice structures, particularly optimized hierarchical lattices, this results in high computational costs. File sizes get very large, making it difficult to perform design modifications, as well as store and export models. Implicit CAD software, such as nTopology, get around this design challenge by representing the intended geometric volume with mathematically defined spatial fields intersected by an isosurface. Thickness can then be applied to the isosurface with signed distance functions or filled to represent a solid. In this way, implicit representations are advantageous for lattice structure design, because the geometries can be rapidly generated and tailored with local optimization, using much less computation resources and file sizes.

An example of the nTopology platform to perform topology optimization was provided by Costa et. al. [18], who used it to design AM components for a hydrofoil wing. This study demonstrated generative design techniques that included field-driven design and lattice structures, in combination with topology optimization to implement variable-thickness shells for improved weight efficiency. While many benefits of the DfAM workflow were highlighted, there were also drawbacks with geometric compatibility when models had to be transferred to other software packages. It reinforced the need for improvement of the AM workflow to eliminate barriers if many of the technology promises are to be realized.

In regards to optimizing the quality of lattice manufacturability, it was found to be ideal to design features to be as thick as possible with high inclination angles relative to the build direction [16]. When lattice geometry is converted into an STL format for manufacturing, file size reduction efforts often result in larger triangulated facets, which reduces the accuracy of the triangular surface mesh representation. An extreme example would be where a circular cross section of a strut is translated into a triangle or square. Therefore, the simplification of surfaces must be carefully inspected to ensure the resulting geometry meets the quality requirements of the manufactured lattice.

It was also reported that local design optimization strategies have been infrequently implemented in AM lattice design, despite the technical advantages [16]. Part of the reason was that designers have to be willing to learn and understand how to implement those opportunities. Typically, lattice architectures are strut-based circular cross-sections that are easy to implement, but do not necessarily provide an optimal response to applied loads. On a large scale, variable cross section beams have been well understood by civil engineers to provide enhanced mechanical properties, but optimized lattice applications that can provide a more uniform stress distribution is only beginning to be applied. Generative design strategies that implement topology optimization are only starting to emerge.

A 2021 review of applications of metal additive manufacturing in aerospace by Blakey-Milner et. al. [19], documented several successful examples, noting the associated benefits and challenges. One example highlighted was a IN718 rocket engine by Cellcore with integrated lattice internal cooling channels. Another was a CubeSat bus structure, shown in Figure 1.6, that was designed by nTopology in collaboration with the U.S. Air Force Institute of Technology. The CubeSat truss structure members were replaced by sandwich beams with a TPMS lattice core, consolidated 150

components into less than 25, and achieved a mass reduction of 50%. Thales Alenia used lattice structures integral to a topology optimized satellite solar panel deployment mechanism, and a satellite “diphasic heat spreader” sandwich panel. For continued aerospace development, the authors stated that the most important areas to focus on are: new alloys, in situ monitoring for flaw detection and digital twin, build risk predictive simulation, non-terrestrial AM, broader implementation of cellular structures (lattices), usage of topology optimization, and multifunctional components.



Figure 1.6. CubeSat bus structure with integral lattice designed by nTopology and the U.S. Air Force Institute of Technology. Credit: nTopology [20].

A 2023 review of AM lattice structures produced from Ti-6Al-4V for bone scaffolds was published by Distefano, Pasta, and Epasto [21]. They defined three classes of lattice structures to be strut-based, skeletal-TPMS, and sheet-TPMS. Of the strut type, they noted that variations of cubic, and the Octet-truss were the most commonly reported on, whereas the Gyroid, Diamond,

and Primitive were the more common TPMS structures, in both sheet and skeletal form. One of the benefits of architected lattices for bone implants was to promote bone ingrowth into the porous titanium structure. However, much debate remains as to the optimum pore size with a wide range of reported values from 100-1000 μm , and a mean value of 522 μm , verified by μCT . The as-built surface roughness of AM lattices can also be a benefit to bone ingrowth up to an R_a of 8.5 μm , although increasing beyond there the permeability, which helps enhance the osteointegration process, begins to decrease.

While Ti-6Al-4V has been demonstrated to be a great biocompatible alloy, its elastic modulus of 110 GPa poses a challenge when placed in contact with Cortical bone (10-20 GPa) or Cancellous bone (0.2-4 GPa) [21]. The ability of a lattice to be designed to reduce the modulus of the parent material and better match that of bone is a key advantage, but which is the most suitable lattice topology is still controversial. Aside from the strut or TPMS geometries, an important lattice parameter is also the relative density, which researchers have tried many combinations ranging from 5% to 90%. This review continued to identify the Gibson-Ashby model as the most relevant and recognizable model for predicting the mechanical properties of a lattice. The collected results showed that lattices with relative densities below 30% fell within the range of cancellous bone elastic modulus. However, the authors stated that further research was needed in the development of a repeatable and robust design methodology for implants.

A review of EB-PBF produced Ti-6Al-4V lattice structures was published by Del Guerclo et. al., in 2020 [22]. They concluded that the EB-PBF process is an excellent technique for fabricating such complex structures that could not be made any other way, with a lower possibility of high- or low-density inclusions. While bulk EB-PBF Ti-6Al-4V microstructures consist of a coarse mixture of α and β grains, higher cooling rates in cellular materials often resulted in a finer

microstructure with the presence of α' martensite. The primary descriptor of periodic lattice geometry was found to be by relative density, and the mechanical behavior was largely investigated experimentally despite the many attempts to simulate it. Failure obtained under compressive loading typically occurred due to brittle fracture, and results were then often used to inform Gibson-Ashby empirical models. Crush bands, that typically occur at 45 degrees at room temperature, tend to occur at random angles when under elevated temperatures. Although, one of the most important roles in the mechanical behavior is the surface roughness, where lower values of R_a and R_z resulted in improved properties, particularly in fatigue conditions.

For the most part, industrial applications of EB-PBF Ti-6Al-4V lattice structures have been investigated mainly for the medical sector due to the higher degree of biocompatibility, and while heat treatment effects on bulk material have been thoroughly analyzed, there is a lack of discussion when it comes to the heat treatment of lattice structures [22]. What was also lacking was any mention of TPMS lattice structures as the authors seemingly discounted their inclusion as a lattice. While not explained, the authors defined that, “a lattice cell consists of a certain number of struts, also called lattice struts, that are joined in one or more nodes.” Thus, such a definition would discriminate against triply periodic minimal surface structures being considered as a “lattice,” despite still representing a non-stochastic cellular material. This suggests some disagreement among researchers as to what topologies constitute a lattice structure.

In a recent review published in 2023, by Changdar et. al. [23], it covered laser AM of stochastic and non-stochastic aluminum cellular materials. They defined two forms of stochastic foams as open cell or closed cell foams, and four types of non-stochastic lattice structures. Under the non-stochastic classification are the traditional 2.5D honeycomb cells, and the 3D lattices of strut-based, shell/plate-based, and TPMS. Prior to 3D printing technologies, stochastic aluminum

foams were produced by a variety of methods such as, injecting molten metal with gas, injecting molten metal with a foaming agent, using a solid-gas eutectic solidification, heat treatment in a mold with metal powder and foaming agent mixture, or heat treatment in a mold with a solidified foam precursor. These methods all resulted in less dense aluminums for light weight applications, but the density and properties were difficult to predict and control. Advances in AM have brought about the ability of complex architected non-stochastic cellular materials to provide this needed precision and design customization.

Among the more commonly reported lattice unit cells for the strut type were variants of the Body-Centered Cubic (BCC) and Face Centered Cubic (FCC), Octahedron, and Kelvin cell; whereas the more common TPMS-based unit cells were the Gyroid and Diamond [23]. Performance of the different lattices has been typically measured empirically through compression experiments, characterized by the relative density. However, the review identified the role of process parameters as being crucial for minimizing defects, which were usually the key drivers for lower than predicted performance. In one example of an Octet lattice from Liu et. al. [24], a finite element model was able to more closely predict the compression performance after imperfections had been statistically distributed throughout.

According to an overview of materials with triply periodic minimal surfaces, published by Han and Che in 2018 [25], the Primitive, Diamond, and Gyroid surfaces, have become some of the most important and commonly observed TPMS structures. Schwarz, in 1865 [26], and later his student Neovius, in 1883 [27], provided the first examples of TPMS's with the Schwarz Hexagonal, Schwarz Crossed Layers of Parallels, Schwarz Primitive, Schwarz Diamond, and Neovius. Schoen, in 1970 [28], described the now famous TPMS Gyroid surface, as well as eleven others. Since then, natural systems like sea urchins, butterfly wings, beetle exoskeletons, and

biological cubic membranes, have been found to display similarities to these complicated and highly symmetrical structures. Artificial systems, such as self-assembled block copolymers, lyotropic and thermotropic liquid crystals, and zeolite crystals, have also been shown to resemble TPMS structures. Equations (1.1), (1.2), and (1.3), describe simple approximations for the Gyroid, Diamond, and Primitive geometries, respectively, that are considered minimal when the threshold value $t = 0$,

$$\sin X \cos Y + \sin Z \cos X + \sin Y \cos Z = t \quad (1.1)$$

$$\cos X \cos Y \cos Z - \sin X \sin Y \sin Z = t \quad (1.2)$$

$$\cos X + \cos Y + \cos Z = t \quad (1.3)$$

where $X = 2\pi x/a$, $Y = 2\pi y/a$, and $Z = 2\pi z/a$, with a as the unit cell parameter.

One of the most recent comprehensive reviews of metal lattice structures mechanical properties was published in 2024 by Miao et. al. [29], which covered 358 literature sources. It covered both strut and TPMS topologies where the most widely investigated TPMS topologies were the Gyroid, Diamond, Primitive, and IWP (Schoen I-graph Wrapped Package), with steel and Ti-6Al-4V being the more common materials. The authors reported one of the existing challenges as being that studies have almost exclusively performed experiments in compression, both static and fatigue, such that there is a need to demonstrate these structures under more complex external loads that better simulate actual working conditions. While manufacturing technologies such as PBF have enabled the ability to produce such structures, specimens still contain many defects that heavily influence the observed behaviors. Additionally, the limitations of FEA to predict the behavior of lattices was also noted, yet it was forecasted that these simulations will become more important in order to efficiently investigate the seemingly infinite lattice structure design space.

Additively manufacturing lattices via PBF processes is not without its challenges, as reported by Echeta et. al. [30], in their 2019 literature review. They point out that these structures are prone defects due to the fact that support structures, which are typically used to mitigate defect formation, cannot be placed internal to the structure because they would not be able to be removed. Therefore, material quality and build accuracy is much lower and more variable for both strut and TPMS type geometries that are not self-supporting with respect to the build direction. High porosity, lack of fusion, cracking, delamination, spatter, and poor surface roughness all occur at higher rates in PBF lattices than solid structures. However, methods for minimizing the defects were reported such as implementing constraints on the design, optimizing process parameters and scan strategies for the geometry, and post processing with chemical etching due to access challenges. While these might sound straight forward, the amount of available literature does not adequately cover the vast combinations of geometries, materials, processes, and parameters, that can all affect the formation of defects in lattice structures.

A review published by Liu et. al. in 2024 focused on factors that affected the mechanical properties of AM Lattice structures [31]. This review largely covered a wide range of strut-based geometries, with very little on TPMS architectures, mainly the Gyroid. Not surprisingly, the performance of lattice structures depends primarily on rational topological design and the material. Most metallic materials that have been investigated have been Ti-6Al-4V and 316L, leaving much room for alternative higher strength materials to be developed. Functionally graded and non-linear gradient structures have been shown to be able to provide tailored properties and much greater energy absorption. However, most studies focused on uniform and linear density distributions, such that there is a lack of complex functionally graded lattice structures in current literature. Quality of the produced material through proper process control, and post processing heat

treatments, has been shown to have large effects on the mechanical properties, yet strategies for improving quality need further research. The predominant applications in the medical and aerospace fields will require the establishment of standards for manufacturing of AM lattices and quality assurance strategies.

The purpose of this work was to review the relevant literature regarding PBF AM, with a focus on the electron beam process, and how it has been applied to complex structures such as lattices. While a few examples of polymer materials are included, the primary materials of interest are metallics, particularly titanium. The alloy Ti-6Al-4V is one of the more popular alloys due to its high strength-to-weight ratio, corrosion resistance, toughness, and challenge to manufacturing via conventional processes. The following chapter will provide an up to date, state-of-the-art look into where this technology research is advancing, as well as the methods and processes used to evaluate the flexural properties of novel lattice sandwich structures manufactured from Ti-6Al-4V via EB-PBF.

Chapter 2. BACKGROUND

There are many classifications of additive manufacturing, which strictly means the buildup of material, as opposed to the removal of it. An example of a traditional subtractive manufacturing method would be the cutting and machining, such as a 5-axis mill or lathe. Currently, the terms 3D printing and additive manufacturing are synonymous with each other, though the latter is most often found in industrialized environments for the production of end use parts. When it comes to the fabrication of metallic components, the most common form of additive manufacturing, is the powder bed fusion (PBF) process. Generally, PBF consists of a thin layer of powder being evenly distributed across an area within a build chamber to form a powder bed. Then a heat source selectively melts a region, fusing the powder in the shape of a cross-sectional slice of the object being manufactured. After that layer completes all designated melting, the build area is lowered the thickness of one layer, and a new layer of metal powder is distributed. The process continues layer-by-layer until all objects within the build volume have been completed, at which point the excess un-melted powder is removed to reveal the final components.

With this method of fabrication, complex metallic structures that could not be manufactured previously can now be realized. In industries where weight is critical to product performance, PBF can produce more weight optimized structures that do not have to conform to all the rules and restrictions of traditional manufacturing methods, e.g. forging, machining, casting, brazing, etc. Porous, or space frame, structures have been common in lightweight engineering by placing material where needed along the most direct load paths. While that has been done at the macro-scale for large structural assemblies, e.g. truss bridges, PBF offers a similar capability at the individual component, or meso-scale with lattice-type structures. The mathematical field of topology optimization, which strives to numerically predict where material should be placed within

a 2D or 3D domain, has grown significantly with the maturation of additive manufacturing into a field often referred to as generative design.

The feedstock materials, most often associated with metal powder bed fusion additive manufacturing, are well-known weldable alloys. Because the process is similar to fusion welding at the micron scale, steel, nickel, aluminum, and titanium alloys, that are commonly welded in industry, work well in PBF processes. For high-temperature, weight-critical structures, titanium is popular for its specific strength, high hardness, and corrosion resistance. Titanium is also difficult to process through traditional manufacturing methods, which makes it expensive, and thus less of an economic hurdle to additive titanium parts.

This chapter will address the state of the art in additive manufacturing powder bed fusion, with a specific focus on research relevant to the utilization of an electron beam (EB-PBF) as the heat source, versus a laser (L-PBF). Then it will go through the latest research in the additive manufacturing of lattice structures, the different lattice types, and applications of topology optimization. The last section will focus on the feedstock material for PBF, primarily with regards to the common titanium alloy Ti-6Al-4V that was used for experimental investigations described in later chapters.

2.1 ELECTRON BEAM POWDER BED FUSION

As the name implies, electron beam powder bed fusion uses a powerful beam of electrons to selectively melt a metal powder feedstock. A key differentiator from L-PBF is that the EB-PBF process must take place in a vacuum environment, where the beam is also used to heat the build chamber to high temperatures, or about half of the material melt temperature ($\sim 0.5 T_m$) [32]. While L-PBF is limited to a single melt pool per laser, a single electron beam emitted from an

electron gun can be split into multiple beams, such that it is capable of maintaining many simultaneous melt pools. An electron beam can also move at speeds many times that of a laser, because it is steered by electro-magnetic fields, rather than a mechanical beam guide like a galvanometer. However, the powder material and other process parameters typically limit the speed of the beam to achieve quality fusion.

The EB-PBF workflow has many similarities with other AM processes, such as starting with a 3D computer aided design (CAD) model of a part ready for manufacturing. That model is then converted into a triangulated mesh that approximates the surfaces of the model, the most common format being STL. The STL format has been the de facto standard for AM files since it was created by 3D Systems when they were first developing stereolithography AM in the mid 1980's [33]. The meshed models are then arranged in a simulated build volume using a commercially available program, the most common of which is called Materialise Magics. This is where the build orientation of the parts and support features are defined as depicted by the third step in the process flow shown in Figure 2.1.



Figure 2.1. EB-PBF process workflow from initial CAD model to fabricated component [34].

Support features in EB-PBF primarily serve as heat sinks to expedite solidification, while the surrounding unmelted powder behaves as an insulator that can trap heat at the build surface. The supports provide a more direct conduction path to the cooler powder layers below in order to avoid distortion of the part due to uneven cooling. Support features are needed on negative, downward facing surfaces that exceed a certain critical angle with respect to the build plane. That critical angle is not fixed, and will depend on the process parameters and the geometry of the part being produced. Supports are sacrificial, and designed to be porous thin-walled structures to facilitate easy removal [35]. Upon completion of the virtual build layout, the entire package of parts and supports is exported to another software package for slicing and beam path generation. Slicing is the process of converting the 3D components into a series of two-dimensional cross-section cuts, based on the defined layer thickness, which are perpendicular to the build direction. Then, depending on each layer's 2D profile melt area and the desired scan strategy, the beam path instructions are generated, and the information is written to a build file to be uploaded to the EB-PBF machine.

2.1.1 *EB-PBF Machine and Process*

The work described in later chapters involved the use of an EB-PBF machine, called the Arcam A2X, which consisted of two main cabinets. The left cabinet contained the controls and high-power electronics, while the build chamber and electron beam column were to the right, as shown in Figure 2.2. The electron beam column is located above the build chamber, and at the top is where electrons are emitted from an electrically heated tungsten filament. It has a current range of 0-50 mA, and 60kV of power acceleration up to 10,000 m/s. There are three electromagnetic lenses along the length of the column that control the beam. An astigmatism lens affects the spot shape, while the focus lens adjusts the spot size (typically a diameter of 250 μm), and a deflection

lens changes the spot position in the XY plane of the build area. The electrons transfer their energy kinetically to the atoms of the powder bed material through elastic and inelastic collisions. A defocused beam is employed during phases of operation when preheating the chamber and sintering the powder bed, whereas a focused beam is applied for the full melting and fusion of the components being fabricated [35],[36],[37].

a)



b)

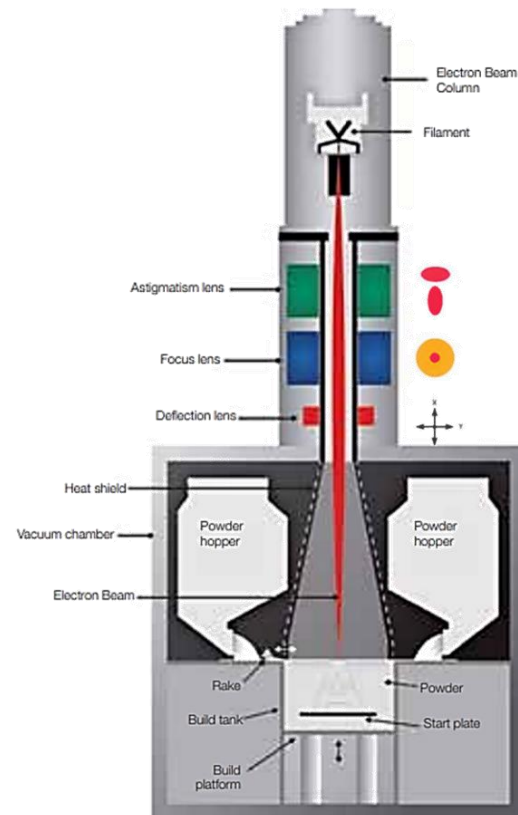


Figure 2.2. a) Arcam A2X EB-PBF system, and b) a schematic of the build chamber with the beam control column on top [36].

Below the electron beam column, in the central region of the build chamber, is the powder bed where the feedstock material is melted. To either side of the build area are the powder hoppers that use gravity to feed powder to the distribution rake. A highly conductive substrate (e.g. stainless steel) build plate is placed on top of a bed of powder to protect the build platform that

controls the layer thickness. The build plate facilitates quick solidification by operating as a heat sink for early layers, and by comparison this plate is rather thin (10 mm) relative to those used in L-PBF systems (50-100 mm), and does not need to be of the same material as the powder due to significantly lower residual stresses. The dissimilar metals between a Ti-6Al-4V feedstock and the steel encourages cracking of the welds between the plate and the parts when the build is later cooled to room temperature. This enables the removal of parts by hand without the need for cutting tools or machining. Because an electron beam will interact with atoms in the atmosphere, it must therefore operate within a high vacuum (normally $1\text{e-}5$ mbar). The vacuum chamber environment is a unique aspect of the EB-PBF process, where the internal pressure is typically regulated to 2×10^{-3} mbar utilizing helium (He). This back pressurization is advantageous in terms of the reduction of potential contaminants, such as oxygen, as well as to suppress the reaction of material particles to electrostatic charge, commonly referred to as powder spreading or smoking [36],[37].

On the other hand, a potential disadvantage of the added He can be the effect of vapor pressure of volatile elements within a pre-alloyed powder that may be at risk of evaporation. For example, with Ti-6Al-4V, the energy required to melt the titanium may vaporize the constituent aluminum, and thus the resulting material will not have the expected proportion of elements and material properties. In a study conducted by Damri et. al. [38], they increased the gas pressure by an order of magnitude, to 10^{-2} , in order to determine if there were reductions to the energy deposition efficiency from the attenuation of electrons. By increasing the He pressure, they sought to improve the thermodynamic stability, and decrease the composition volatility of pre-alloyed solids that contain high vapor pressure components. The authors employed a novel Monte Carlo code to calculate the energy deposition of electrons, and correlated the simulations with temperature increases measured in thermal experiments. They found that there was only a 1.5% reduction in

energy deposition efficiency, indicating that increasing the gas pressure is a viable method to reducing the risk of alloy element evaporation. While that could be a method of expanding the list of materials the EB-PBF process supports, He gas is also a consumable, and thus increasing its consumption may negatively impact the economics of the process.

Once the chamber is at the desired vacuum level, and the rest of the Arcam A2X setup procedure is complete, the actual printing process begins with a preheat phase to bring the build plate up to the desired temperature (e.g. 730 °C for Ti-6Al-4V). This initial preheating step takes place with a high velocity (35 m/s), high current (48 mA), defocused beam scan. When the desired temperature is reached, as detected by a thermocouple beneath the build plate, the rake fetches powder from one of the hoppers and spreads it across the build area to make a smooth powder bed. Every layer also starts with a preheat phase designed to sinter the powder to prevent it from reacting kinetically to the negative charge of the electron beam. Additionally, preheating the powder bed helps to maintain an overall constant energy input per layer.

There are two steps to the layer Preheat phase. Preheat 1 scans the entire powder bed with a high beam current to raise the newly distributed powder temperature up to a “jump-safe” level (about $0.5 T_m$ Kelvin). Then Preheat 2, using a similarly high beam current, subsequently rescans an area circumventing the regions to be melted with a defined offset (e.g. 4 mm); the second preheat step further raises the local powder temperature to a “melt-safe” level (about $0.8 T_m$ Kelvin) [32]. After both preheat steps have sintered the powder bed, a focused beam first melts the contours of the parts, followed by melting of the part infill, called a hatching scan. Depending on the time spent performing the contour or hatching scans, a post-melt heating phase will be performed using the same parameters as Preheat 2. The objective of the post melt heating is to counteract excessive cooling due to conduction or radiation that occurs during the contour and

hatching phases. When the layer process shown in Figure 2.3 is complete, the build platform is lowered one layer thickness (e.g. 50 μm), and then the rake fetches more powder from the flanking hoppers to distribute it evenly across the build area to form the next layer.

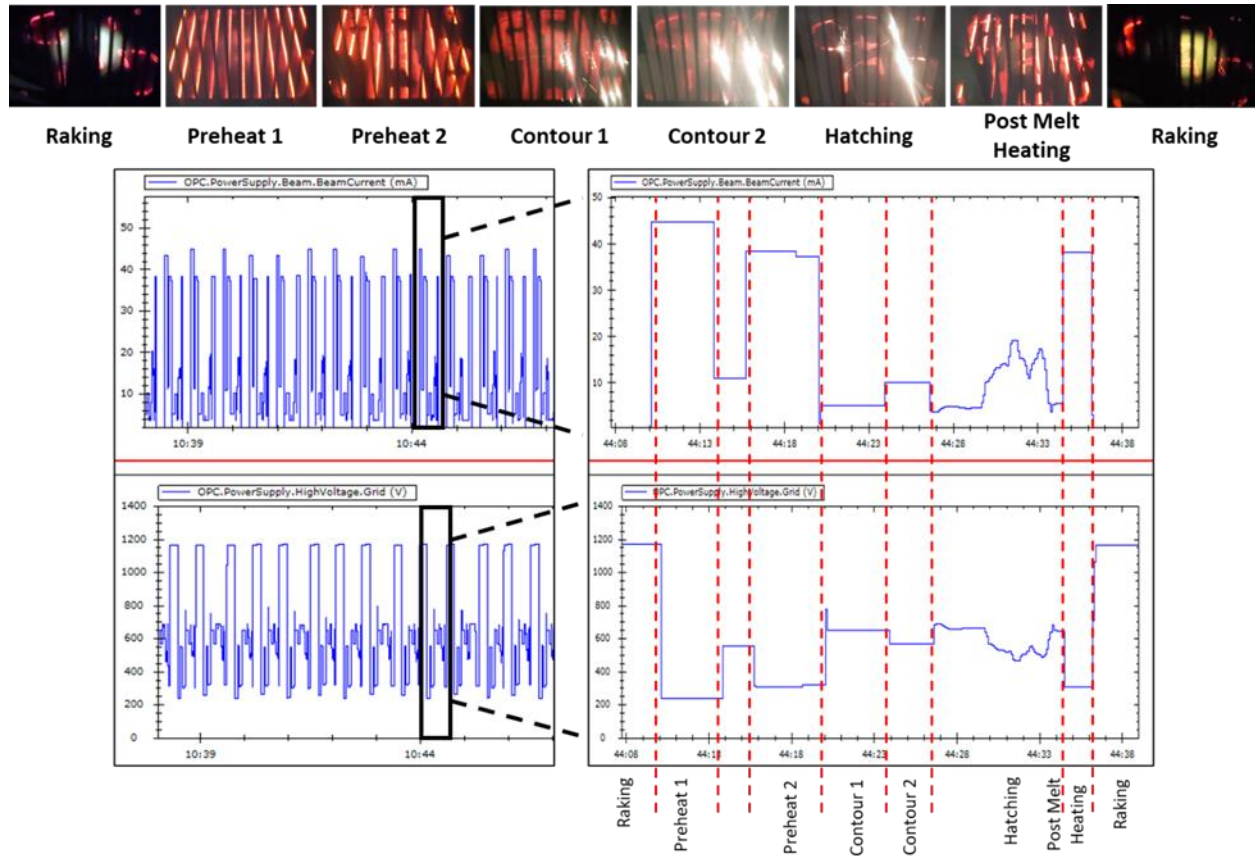


Figure 2.3. EB-PBF build layer process broken down into each phase with the beam current profiles displayed in the top graph and the grid voltage displayed below [34].

A build is complete when the last layer is melted that contained a slice of the objects being produced. After completion, the sintered powder cake containing the melted parts must cool before it is extracted and placed in the Arcam Powder Recovery System (PRS) for post processing. This step is similar to sand or grit blasting, but utilizes the same powder as the build material for the media, to separate the parts from the used powder that will be filtered and reclaimed. The EB-PBF process can accommodate a wide variety of metal alloys due the high energy density of the

electron beam and is thus also suitable for refractory and reactive materials [39]. Typically, the EB-PBF process employs a powder size distribution (PSD) of 45-106 μm in diameter [36], and it can accommodate not only virgin material, but also reused powder. Reused powder helps to improve the economics of the process by minimizing waste; however, irregular oxygenated particles can impact print quality, mechanical properties, and material microstructure [40].

Studies into the EB-PBF process physics are ongoing, as demonstrated by Riedlbauer et. al. [41], whom investigated the lifetime and width of the melt pool with Ti-6Al-4V. They simulated the process with a finite element model (FEM), shown in Figure 2.4, using a homogenized approach to the unmelted powder layer, and a separate continuum for the molten and re-solidified material. The thermal model they implemented involved the transient heat transfer equation and Fourier's law to calculate the heat flux, where the electron beam power input was accounted for in the volumetric heat source term. The electron beam was modeled as a moving heat source while taking into account absorption, penetration depth, electron backscattering, and transmission.

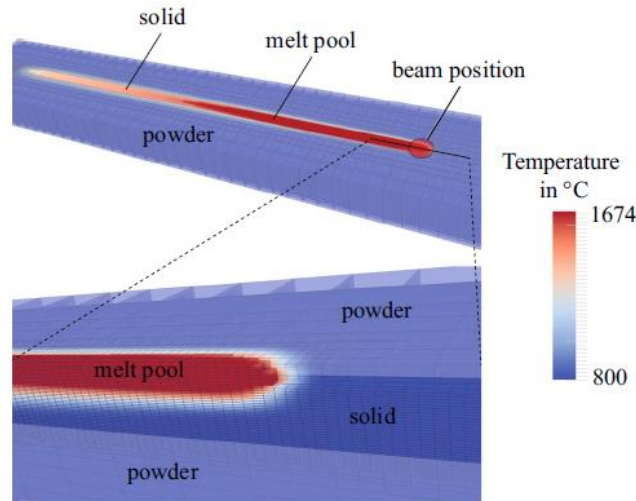


Figure 2.4. Simulation of the EB-PBF process (800 W at 4 m/s), with adaptive mesh refinement in the vicinity of the beam [41].

Their material behavior model for Ti-6Al-4V considered the different phases by temperature from the thermal model, which used a melt temperature of 1674 °C, and a solidification temperature of 1615 °C, with density varying based on the material phase, $\rho_{\text{powder}} = 2564 \text{ kg/m}^3$, $\rho_{\text{melt}} = 3800 \text{ kg/m}^3$, $\rho_{\text{powder}} = 4420 \text{ kg/m}^3$. Initial temperature of the simulation space was assumed to be 800 °C everywhere, while beam spot size was set to 400 μm . The numerical calculations of the FEM were solved first through temporal discretization with a time step that ranged from 4-20 μs , and then spatial discretization with a fine adaptive mesh [41].

In order to compare their model with experimental measurements, the melt pool lifetime was captured through the evaluation of highspeed camera images at a frame rate of 10,000 fps, and melt pool width was approximated by the wall thickness of the as-built part after cool down. The simulation and experimental results for melt pool lifetime were in good agreement, ranging from 1-4 ms as line energy increased from 0.1-0.4 J/mm. Experimental wall thickness measured around 475 μm with a standard deviation of roughly 100 μm , although the simulation tended to predict melt pool widths from 350-475 μm depending on the line energy. The difference was likely due to the difficulty in measuring the resulting chaotic melt pool following solidification, though overall the simulation compared well in relation to changing line energies [41].

2.1.2 *Electron Beam Powder Bed Charging*

In EB-PBF, the electron beam is the sole energy source used for preheating, sintering, and fusing the powder particles. The interaction of high energy electrons with metallic powder creates a charge distribution throughout the build chamber. However, when the electrostatic charge surpasses a critical limit, the negatively-charged particles experience a repulsive force that results in powder motion. This effect cascades almost instantaneously sending powder in all directions

throughout the chamber, and the repulsive forces are even strong enough to overcome the force of gravity. The result is a powder cloud where powder is sent vertically about 1 meter to the top of the electron beam column, and hence referred to in industry as “smoking.” The phenomenon only exists as long as the power beam remains on, but once it is shut off either manually or automatically by an arc trip fault, the powder returns to a stationary position.

Smoking is typically suppressed through a number of mitigation steps that have been developed over the years, largely through trial-and error. One step is by increasing the particle size such that it requires a higher repulsive force to cause movement in opposition to the force of gravity. Another involves the elevated build chamber temperature, along with the Preheat 1 and Preheat 2 phases of each layer. Preheat 1 initially secures the particles with a light sinter, making the powder bed “jump-safe,” before Preheat 2 scans the local areas to be fully melted, making the particles “melt-safe” by lowering the difference in thermal conductivity between components and powder. As described previously, there is also a small partial pressure of helium that is injected into the vacuum system that aids in reducing the risk of smoking. The helium, being an inert gas, is introduced during the melting phase and forms He^+ ions above the powder bed to pick-up excess electrons [42], [43], [44], [45].

From 2009, Eschey, Lutzmann, and Zaeh [42], wrote about increasing the process stability of EB-PBF through experiments in process parameters. They experienced powder smoking for a range of particle shapes, preheat temperatures, beam currents, accelerating voltage, and beam spot diameter. Their layer thickness was 100 μm with a particle size range from 10-60 μm . It was noted that the powder always smoked when using spherical particles, particularly during preheating. However, smoking did not occur with irregularly shaped particles when they were preheated, but they always smoked at room temperature. This work pointed to higher process

temperatures providing better powder stability, while also indicating that spherical particles are more susceptible to smoking.

Some researchers have attempted to develop models of the behavior in order to predict if smoking will occur. In the work published by Cordero et. al. [43], they used theory and experiments to investigate the causes of powder bed charging to formulate an analytical model. The model was based on a “core-shell” particle structure since they found through experimentation that there is commonly an electrically insulating oxide overlayer, likely composed of Al_2O_3 and TiO_2 , covering gas-atomized Ti-6Al-4V powder in the size range of 40-103 μm . Their model showed that uncharged particles develop an electric dipole moment from charged particles in their vicinity, and that they can still experience electrostatic forces far removed from the beam. This explains why smoking can involve a region of powder much larger than the beam diameter.

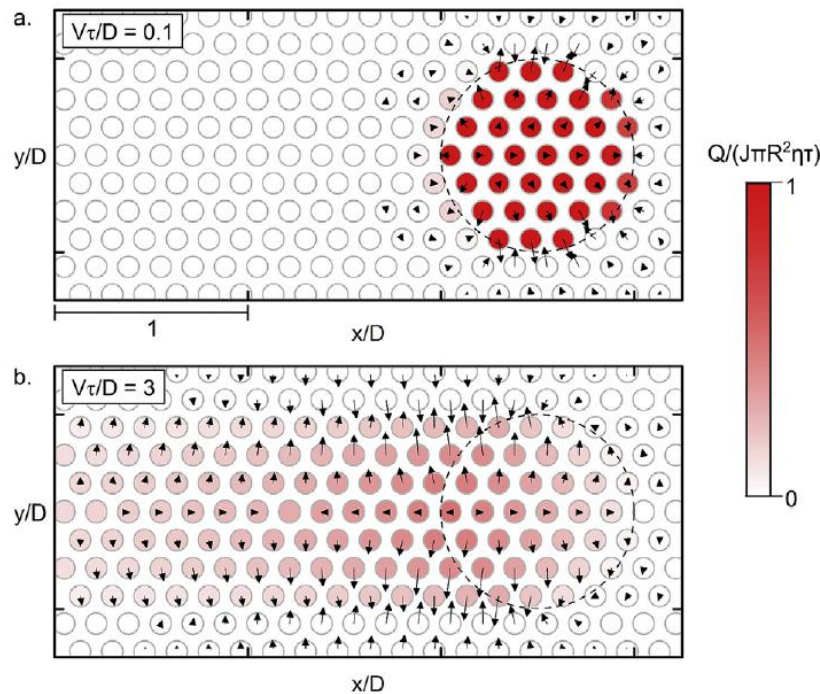


Figure 2.5. Distribution of charge in a single layer of monodisperse particles irradiated by a circular beam (dashed circle). Particle color corresponds to the net charge, and arrows indicate the direction and relative magnitudes of the electrostatic force [43].

The analytical model was developed by first expressing in Equation (2.1) and (2.2) the approximate charge, Q , on a particle as a function of time, $I(t)$, that described the electric current injected into the particle. The charge dissipation time constant, τ , assumes an ohmic resistor, and equals the product of permittivity and resistivity of the oxide overlayer which, depending on temperature, microstructure, and composition of overlayer, can range from picoseconds to days. The particle current was assumed to have a uniform current density, J , for a time, Δt , beginning at $t = 0$, and the beam is collimated to completely span the particle radius, R , with an absorption efficiency, η [43].

$$\frac{dQ}{dt} + \frac{Q}{\tau} = I(t), \quad Q(0) = Q_0 \quad (2.1)$$

$$I(t) = \begin{cases} J\pi R^2 \eta, & 0 \leq t \leq \Delta t \\ 0, & \text{otherwise} \end{cases} \quad (2.2)$$

Equation (2.3) shows the result after solving for $Q(t)$ that assumes spherical particles that are monodisperse and appreciably smaller in diameter compared to the beam. For a beam that travels along the x-axis in the positive x-direction, $t = 0$ is when the leading edge of the beam passes through the origin. A particle at the left edge would follow Equation (2.4), where $\Delta t = \frac{kD}{v}$ ranges from 0.01 - 1 ms, and is a function of the beam diameter, D (0.2 - 2 mm), beam velocity, v (0.5 – 10 m/s), and a constant k with an order of unity to account for the position of the particle relative to the center of the beam [43].

$$Q(t) = \begin{cases} 0, & t < 0 \\ J\pi R^2 \eta \tau \left(1 - e^{-\frac{t}{\tau}}\right), & 0 \leq t \leq \Delta t \\ J\pi R^2 \eta \tau \left(1 - e^{-\frac{\Delta t}{\tau}}\right) e^{-\frac{t}{\tau}}, & \Delta t \leq t \end{cases} \quad (2.3)$$

$$I(t) = \begin{cases} J\pi R^2 \eta, & \frac{x}{v} \leq t \leq \frac{x}{v} + \Delta t \\ 0, & \text{otherwise} \end{cases} \quad (2.4)$$

Therefore, the charge distribution for a row of irradiated particles during the melt operation is reflected by Equation (2.5). This calculates the charge distribution in the top-most layer of particles when combined with a function that also describes the shape of the beam. It reduces to Equation (2.6) for the maximum charge that would increase as the dimensionless group $\frac{v\tau}{D}$ decreases. If that ratio is large, then the particles will retain their charge long after the beam passes over which is undesirable [43].

$$Q(t) = \begin{cases} 0, & t < \frac{x}{v} \\ J\pi R^2 \eta \tau \left(1 - e^{-\frac{t}{\tau} + \frac{x}{v\tau}}\right), & \frac{x}{v} \leq t \leq \frac{x}{v} + \frac{kD}{v} \\ J\pi R^2 \eta \tau \left(1 - e^{-\frac{kD}{v\tau}}\right) e^{-\frac{t}{\tau} + \frac{x}{v\tau}}, & \frac{x}{v} + \frac{kD}{v} \leq t \end{cases} \quad (2.5)$$

$$Q_{\max} = J\pi R^2 \eta \tau \left(1 - e^{-\frac{kD}{v\tau}}\right) \quad (2.6)$$

As described in Section 2.1.1, during the Preheat 1 operation, the beam will raster across the same particle multiple times so that $I(t)$ is actually periodic, with a period $T = \frac{L}{v}$ for the time it takes to raster across the preheat area, and L is the product of the raster length and the number of hatches in a preheated area. In the preheat phase, there are three important time scales τ , Δt , and T , that are related with charge dissipation, beam-particle interaction, and rastering processes. Their relative magnitudes largely determine the amount of charge that accumulates, where $\frac{\Delta t}{T} \ll 1$ because the beam diameter $D \ll L$, and $\frac{\tau}{T}$ and $\frac{\tau}{\Delta t}$ can be greater or less than 1 depending on τ . In the case that $\frac{\tau}{T} < 1$, then the charge will completely dissipate before the next pass, similar to melting, but if $\frac{\tau}{T}$ is large, the charge will accumulate with each pass of the beam [43].

The model development continued by approximating $I(t)$ as a series of delta functions shown in Equation (2.7), that has a magnitude of $J\pi R^2\eta\Delta t$, a period of T , and where n is an integer. The maximum charge then goes to Equation (2.8), with Equation (2.9) describing the full spectrum of preheating charging behavior. When the period is less than the charge dissipation time constant, $T < \tau$, then the charge is unable to dissipate between beam passes and $Q_{\max}$ increases with τ . Changes in τ can cause $Q_{\max}$ to vary over orders of magnitude, but $Q_{\max}$ is limited by either the onset of smoking or melting [43].

$$I(t) = J\pi R^2\eta\Delta t\sum_n\delta(t - nT) \quad (2.7)$$

$$Q_{\max} = \sum_{n=0}^{\infty} J\pi R^2\eta\Delta t e^{-\frac{nT}{\tau}} = J\pi R^2\eta\Delta t \left(\frac{1}{1 - e^{-\frac{T}{\tau}}} \right) \quad (2.8)$$

$$Q_{\max} = J\pi R^2\eta\tau \left(\frac{1 - e^{-\frac{\Delta t}{\tau}}}{1 - e^{-\frac{T}{\tau}}} \right) \quad (2.9)$$

During preheating, the powder bed maximum electrostatic force, F_Q , is described by Equation (2.10), which is greater between charged and uncharged particles. It takes into consideration the vacuum permittivity ϵ_0 , as well as the ratio of oxide thickness to particle radius $\frac{h}{R}$ that is on the order of $\sim 10^{-4}$. Smoking should only occur when F_Q is much larger than gravity, so that can be represented by the inequality of Equation (2.11) where g is the acceleration due to gravity and ρ is the density of the feedstock material [43].

$$F_Q = \frac{\pi R^2}{16\epsilon_0} \left(J\eta\tau \left(\frac{1 - e^{-\frac{\Delta t}{\tau}}}{1 - e^{-\frac{T}{\tau}}} \right) \right)^2 f\left(\frac{h}{R}\right) \quad (2.10)$$

$$\rho g \frac{4}{3} \pi R^3 < \frac{\pi R^2}{16 \epsilon_0} \left(J \eta \tau \left(\frac{1 - e^{-\frac{4t}{\tau}}}{1 - e^{-\frac{T}{\tau}}} \right) \right)^2 f\left(\frac{h}{R}\right) \quad (2.11)$$

For typical EB-PBF Ti-6Al-4V particles, the left side of Equation (2.11) is on the order of 10 nN; therefore, the charge F_Q on the particles must be kept lower than 10^{-14} C. By using values representative of the EB-PBF preheating process, it was shown that $\tau < 0.5$ ns to avoid powder smoking. The model also shows that smoking would be unavoidable if not for gravity, adhesion, and friction, with gravity being the largest stabilizing force by orders of magnitude. Strategies for suppressing smoking, revealed by the right-hand side of Equation (2.11), included increasing the beam speed, decreasing the beam diameter, decreasing the current density, increasing the particle size, or decreasing τ by decreasing the resistivity and permittivity of the oxide overlayer. The resistivity can be reduced through increasing the temperature of the build envelope, as well as through other processes to prevent oxide growth [43].

Additionally, the argument that the particles melt before they smoke was modeled by incorporating a criterion for the critical current density associated with melting, J_{melt} in Equation (2.12), and the criteria for smoking, J_{smoke} from Equation (2.11), and establishing the inequality in Equation (2.13). This incorporates additional process variables such as the electron beam acceleration voltage V_{acc} , the workpiece thermal conductivity λ , the specific heat per unit volume c , and the melting and preheat temperatures T_m and T_0 , respectively. If the inequality in Equation (2.13) is satisfied then it indicates that the particles will melt before they smoke [43].

$$J_{\text{melt}} = \left[\frac{c \lambda \pi}{2} \left(\frac{(T_m - T_0)}{\eta} \right)^2 \frac{v}{V_{\text{acc}}^2 D} \right]^{1/2} \quad (2.12)$$

$$\frac{128}{3\pi} \frac{D\varepsilon_0 R \rho g}{c\lambda v} \left(\frac{V_{acc}}{T_m - T_0} \right)^2 > f\left(\frac{h}{R}\right) \left(J\eta\tau \left(\frac{1 - e^{-\frac{\Delta t}{\tau}}}{1 - e^{-\frac{T}{\tau}}} \right) \right)^2 \quad (2.13)$$

When EB-PBF representative values are input into Equation (2.13), the inequality is satisfied and the model predicts that if the powder did not smoke during preheating, then it will not smoke during melting. Overall, the model is consistent with trends observed experimentally, where larger particle sizes, higher powder bed temperatures, and higher beam speeds help suppress smoking, and indicates that smoking is more likely during the preheating phase, than during melting [43].

Shanbhag and Vlasea [44], examined how the EB-PBF preheat temperature affects Ti-6Al-4V powder sintered neck size in an Arcam A2X. They found that the thermal conductivity of the powder cake is dependent upon the packing density, contact size ratio, and coordination number, and then derived an empirical model. It was concluded that reducing the preheat temperature from 730 °C to 650 °C resulted in only a 3% change in layer-wise density and a 12% change in the effective thermal conductivity. This research provided better comprehension of the layer-wise preheating phases in EB-PBF and how much room there is for process optimization.

In a study conducted by Landau, et. al. [32], the authors aimed to characterize the thermal boundary conditions of the EB-PBF Arcam Q20+ build chamber such that they could build a full-scale finite element thermal model. This would enable the evaluation of part positioning consequences in terms of the different preheat temperatures for “jump-safe” and “melt-safe” process phases. The simulation shown in Figure 2.6 combined heat conduction with open cavity radiation in a 3D model, and then it was validated experimentally with thermocouples installed on the inter-chamber heat shields and build plate. This was important to understand the initial preheat phase since the smoking phenomenon is sensitive to preheat temperature.

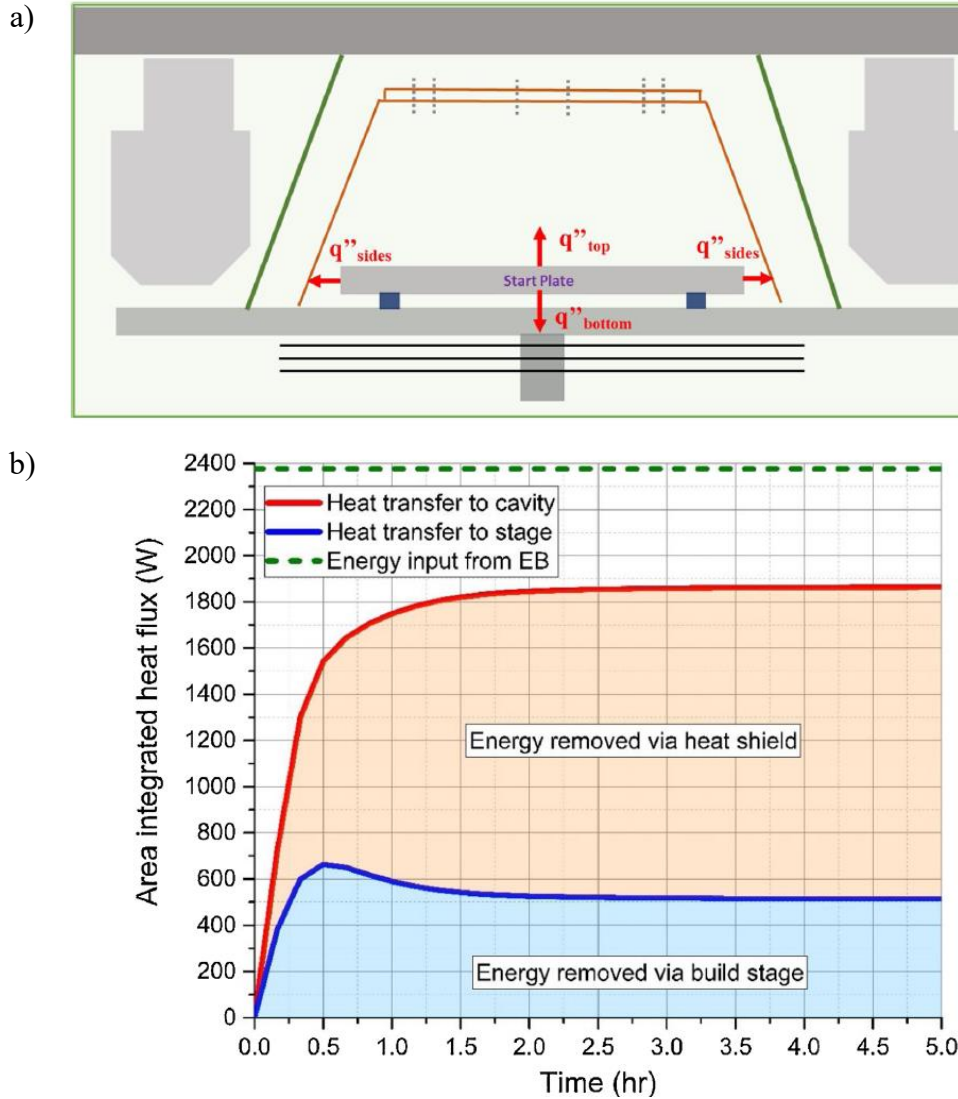


Figure 2.6. Simulation results of EB-PBF build plate energy balance results: a) schematic of build chamber with heat flux directions, b) area integrated heat flux over build plate surfaces [32].

When looking into the thickness dependency of mechanical properties for thin-walled titanium parts produced via EB-PBF, Algardh, et. al. [46], used Ti-6Al-4V powder feedstock with a size distribution of 25-45 μm , rather than the 45-105 μm normally used with EB-PBF. They described how the finer powder size distribution necessitated significant deviations from the standard commercially available process parameters in order to avoid powder charging that

resulted in smoke events. During the preheating phase, the beam focus was decreased from 35 mA to 10 mA, and the scanning speed was increased from 14.7 m/s to 22 m/s by ramping the current from 5 mA to 25 mA. Decreasing the focus meant spreading the beam intensity and charge over a larger area. The scan repetitions were also increased to 15 to maintain a similar energy input and sustain sufficient preheating. These changes align with the sensitivities seen in analytical models where the particle size distribution (PSD) greatly affects the sensitivity of powder to smoking in EB-PBF. While the authors adjustments helped to reduce the charge accumulation rate in the particles, they did not report any temperature effects, or whether the preheat rate changed when using these alternate process parameters.

2.2 LATTICE STRUCTURES

Light weight metal foams, also known as cellular metals, are a solid material with deliberate voids within their volume. Typically, they are very difficult to produce through conventional methods, but the more complexity agnostic powder bed fusion manufacturing processes have made these cellular metals a recent area of interest. Metal foams can be broken into two topological categories as seen in Figure 2.7, stochastic, which has random variations in cell shape and size and are more commonly associated with the term “metal foam,” or non-stochastic (i.e. architected materials), which have repeating or periodic structures [23], [47]. These materials have been investigated for use in a number industries for applications involving multi-function structure, thermal management (heat exchangers, insulation, controlled CTE, light-weighting, vibration dampening, acoustic attenuation, and energy absorption. However, much of the research for additive lattice structures has been focused around the application of medical implants to get non-reactive metals like Ti-6Al-4V to match the modulus of different types of bone [35], [47], [48], [49], [50], [51].

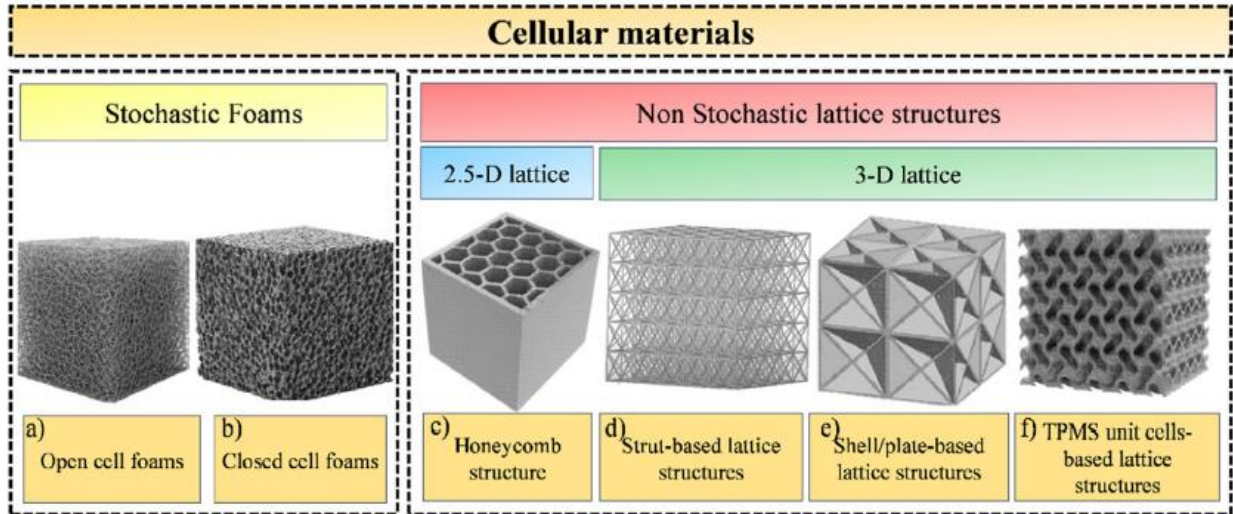


Figure 2.7. Cellular materials overview of stochastic and non-stochastic lattice structures with examples of, a) open-cell foam, b) closed cell foam, c) honeycomb, d) strut-based lattice, e) plate-based lattice, and f) triply periodic minimal surface (TPMS)-based lattice [23].

2.2.1 Uniform Lattices

The difficulty in producing cellular metals has not only been in manufacturing, but also design. CAD modelling of lattice structure topology is computationally and labor intensive, which to date has heavily limited the kinds of geometry to be produced by additive manufacturing. Early investigations involved strut type lattice topologies that were often modeled after crystalline atomic structures, such as the Body-Centered Cubic (BCC), Face Centered Cubic (FCC), or Diamond [52], and these subsequently expanded into more complicated Hexagonal Close Packed (HCP) or Dodecahedron unit cells [53]. In strut-type lattice structures, repeating nodes are uniformly spaced around a volume, with struts connecting at the nodes, and where the struts either have circular or rectangular cross sections. A few studies have investigated graded lattices, topologies that vary in density and/or cell size, and multi-morphology where the lattice transitions from one topology to another [51], [54], [55]. More recently, an intriguing mathematically derived

topology called a triply periodic minimal surface (TPMS), as shown in Figure 2.8, has been the focus of studies that also involve metal additive manufacturing.

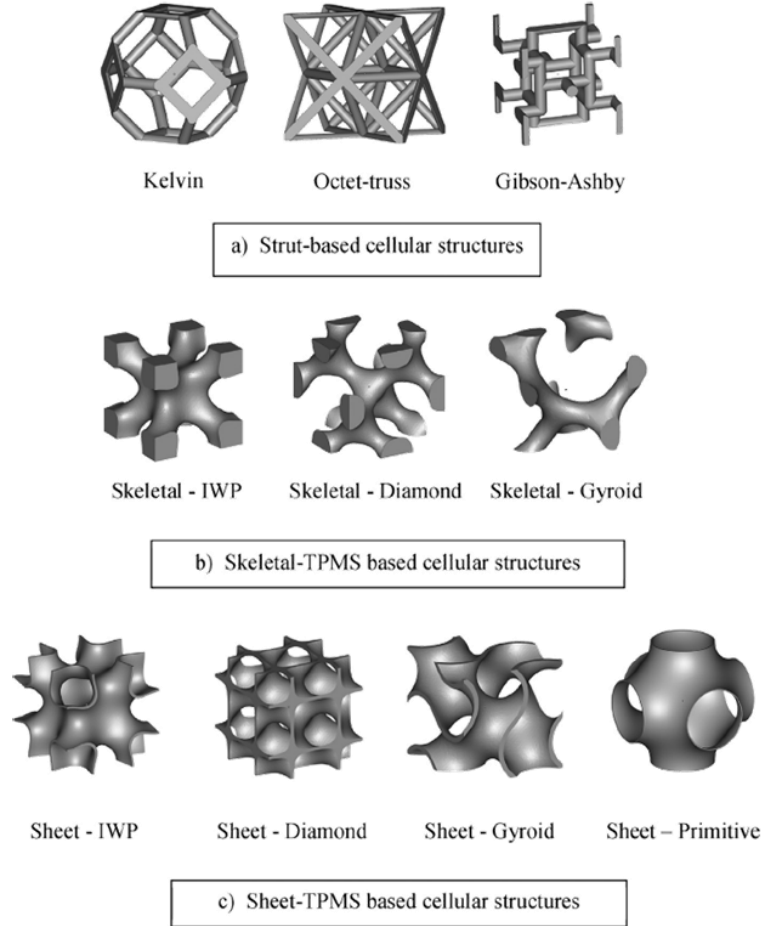


Figure 2.8. Depictions of Strut and TPMS type lattices, with TPMS broken into skeletal and sheet structures [56].

The topological process of designing a lattice revolves around the relative density, or volume fraction, which is simply the ratio of the density of the lattice to the density of the base solid ($\rho_{RD} = \frac{\rho_{\text{lattice}}}{\rho_{\text{solid}}}$) [50]. The relative density is controlled by the lattice thickness t and the unit cell size l , such that the thickness to unit cell size ratio is (t/l) [57]. If the unit cell size (l) is fixed, then the relative density is solely controlled by the thickness (t). To maintain the same relative density as

l decreases, then t must decrease as well. Therefore, the minimum relative density limit, with a fixed l , is the minimum possible wall thickness that can be achieved by the additive manufacturing process.

In PBF, the minimum thickness or feature size is primarily driven by the beam melt volume, and the powder particle size. However, many studies have shown that the relative density of the additively manufactured lattice is generally larger than the design and was demonstrated with both EB-PBF and L-PBF Ti-6Al-4V, as well as Maraging Steel. This is due to the interaction between the melted volume and the powder bed where excess material is partially melted and merged with the part, increasing the surface roughness and solid volume. As the surface area of the lattice increased, the deviation from the design relative density also increased.

Some of the common issues with the metal additive manufacturing of lattice structures includes size constraints, powder removal, surface roughness, and layer-to-layer lattice connectivity. Many published works have been focused on pushing up against the minimum producible size constraint, with design thicknesses under 1 mm. While the beam spot size with EB-PBF is small (e.g. 250 μm), the width of the melt pool is much larger (600-700 μm), plus any additional powder that agglomerates around the edges limits the minimum feature size [34].

Powder removal in EB-PBF is more difficult than L-PBF because the powder bed has to be sintered at every layer to prevent the powder particles from scattering when hit with the large negative charge of the electron beam. The sintered powder must be removed from the printed components by blasting it with compressed air mixed with titanium powder. This method largely limits powder removal to line-of-sight, and access becomes challenging the smaller the lattice unit cells are. It was reported that powder could not be removed from pore sizes below 800 μm [50]. The surface roughness is a larger issue with small lattices because the surface variation is on the

same order of magnitude as minimal thickness geometries. Due to the layer-by-layer nature of the manufacturing process, shallow lattice angles relative to the build direction combined with minimal thicknesses, result in poor overlaps and transitions, as well as poor strength and stiffness [47].

One of the earliest publications of Ti-6Al-4V lattices produced by EB-PBF was submitted in 2009 by Murr et. al. [58]; however, rather than the periodic structures of what is typically thought of as lattices today, open cellular stochastic foams were produced for characterization via optical and SEM imaging. In order to produce CAD models at the time, the authors relied on CT scans of aluminum alloy foams which were much easier to produce due to their lower melting temperatures and being less reactive to mold materials. The relative densities that were produced ranged from 0.11 to 0.45, with cell wall thicknesses from 0.4-1.9 mm, and elastic moduli was measured non-destructively using a resonant frequency and damping analyzer (RFDA) according to Equation (2.14),

$$E = \zeta m f_r^2 \quad (2.14)$$

where ζ is the specimen geometrical shape factor, m is the mass of the specimen, and f_r is the resonant frequency. The measured values were compared with the Gibson-Ashby model and found to be consistent, as stiffness varied inversely with porosity.

Relative density (i.e. porosity, or volume fraction) is the most important feature of metallic foams because mechanical properties increase with higher relative density. The Gibson-Ashby analytical model was originally developed for open-celled foam materials with randomly distributed (i.e. stochastic) porosity and perfectly smooth surfaces [59]. It aims to describe the porous material elastic modulus Equation (2.15), or yield strength Equation (2.16), as it is related with the relative density [48], [56], [60].

$$\frac{E}{E_0} = C_1 \left(\frac{\rho}{\rho_0} \right)^{n_1} \quad (2.15)$$

$$\frac{\sigma}{\sigma_0} = C_2 \left(\frac{\rho}{\rho_0} \right)^{n_2} \quad (2.16)$$

The constants C and n in the equations are determined experimentally and will vary depending on the architecture of the unit cell and the manufacturing method.

Ali, Abdi, and Sun [61], investigated the elastic properties of various strut-based lattice structures through compression tests and compared them to numerically calculated properties. Their study used polymer SLS and DLP produced samples for experiments, which demonstrated how the performance of a lattice is not just heavily dependent upon the design of the unit-cell, but also the process used for fabrication and its material. For the SLS samples, they were fabricated from Nylon PA12, whereas DLP uses a cured photopolymer resin. The Gibson-Ashby power-law model was employed as a framework for comparing the stiffness of the Body-Centered-Cubic (BCC), Face-Centered-Cubic (FCC), and Octet lattice structures for different relative densities ranging from 10-60%. From their results it was determined that low relative densities aligned with the Gibson-Ashby model, but that beyond 30% a second-degree polynomial model provided a better fit.

In a study conducted by Hacisalihoğlu [62], five different strut type lattices based on the BCC, Diamond, and FCC geometries were fabricated via L-PBF from Ti-6Al-4V. The levels of porosity produced for each configuration were 60%, 75%, and 90%, and the specimens were subsequently compression tested to compare with numerical finite element analysis. The unit cell edge lengths were very small, with the smallest for the higher density being 1.6 mm and the largest for the lowest density being 4.55 mm. The lowest density samples showed a majority of the deformation

at the boundaries, while the higher density specimens deformed more uniformly with improved stress distribution.

Babamiri et. al. [63], examined the interplay of the structural topology and the microstructural properties of AM lattice structures where they introduced a design optimization approach to achieve a particular macroscopic mechanical response. At the unit cell level for Octet and Rhombic Dodecahedron truss topologies, the deformation mechanisms were translated to the global strain in FEA simulations. The unit cell size and relative density were constant at 4 mm and 30%, while the strut diameters were varied depending on orientation and ranged from 0.424 mm to 0.766 mm. The new topologies were manufactured from L-PBF Inconel 718, and after compression tests, they showed a 40% improvement in energy absorption capacity, and a 72% improvement in flow stress.

Traxel et. al. [64], used the L-PBF process to produce various lattice structures from Ti-6Al-4V. Their lattice unit cell designs were based on the common strut type natural crystal structures HCP and BCC, and they created unique combinations of both. The designed relative densities were not very consistent and varied from 34.7-47.1%, with very small unit cell sizes from 1.76 mm to 3 mm, and strut diameters on the order of 0.5-0.7 mm. After the lattice structures were experimentally tested in compression, the mixed BCC/HCP structures outperformed the uniform topologies in compressive strength, while the HCP demonstrated a higher resistance to crack propagation than the BCC. Not surprisingly, the higher relative density structures did better so the lack of control over this parameter makes the results difficult to directly compare. However, this work did demonstrate the ability to tailor a lattice structure with multiple topologies to get a wide range of strengths (Modulus: 15.1-76.5 GPa, C-YS: 225-511 MPa, C-US: 301-663 MPa).

Because strut type lattice structures are often modeled after crystalline atomic structures, Liu et. al. mimicked a technique of strengthening crystalline structures at the micro-scale by adding spheres at the mid-point (edge center interstitial lattice), or at the nodes (vertex node substitutional lattice) of the macro-scale lattices, as shown in Figure 2.9 [65]. This approach is similar to a material optimization where mid-span spheres are attempting to limit the effect of buckling failure modes, while node spheres would serve to reduce the high stress concentrations at those locations. They designed cubes of 5 unit cells in X, Y, Z, with a constant strut length of 5 mm, diameter of 0.25 mm, and variable spherical radii that ranged from 0.5 mm to 1.0 mm. Their specimens were manufactured via L-PBF from Al-4.2Mg-0.6Sc-0.2Zr, and noted that an island scan strategy was employed.

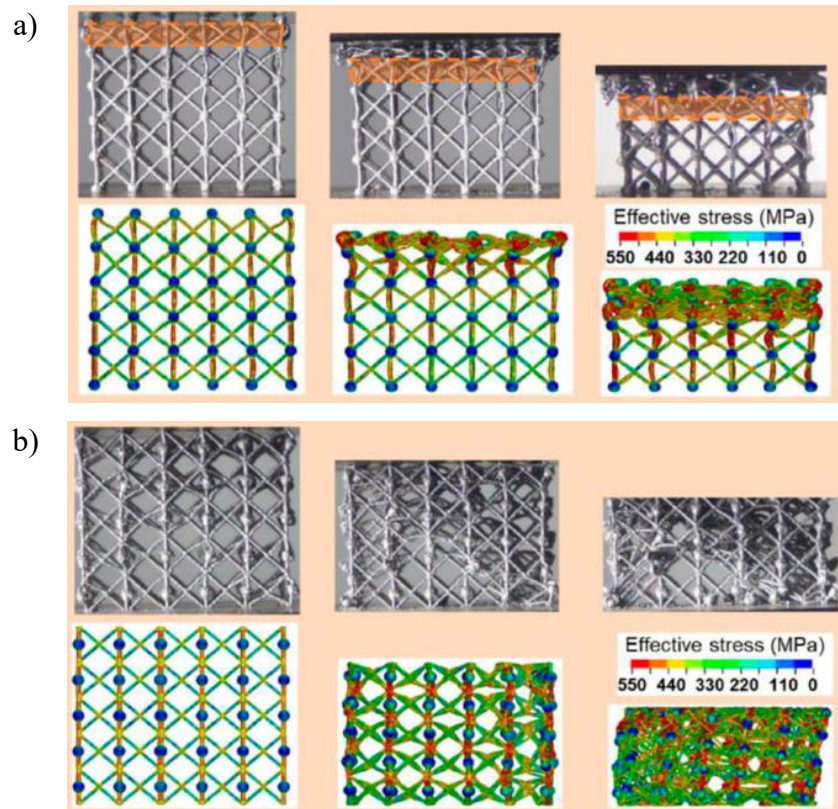


Figure 2.9. Experimental and simulation deformations at 8%, 20%, and 40% applied strain for a) vertex node substitutional lattice, and b) edge center interstitial lattice [65].

The authors also built a numerical model where the deformation gradient F is based on the first Piola-Kirchhoff stress tensor P , Equation (2.17), such that the Cauchy stress is given by Equations (2.18) and (2.19) with the second Piola-Kirchhoff stress S .

$$F = F^e F^P \text{ or } P = \det(F) \sigma F^{-T} \quad (2.17)$$

$$\sigma = \frac{1}{\det(F^e)} F^e S F^{eT} \quad (2.18)$$

$$S = \mathbb{C}:E^e \quad (2.19)$$

The as-built structures were μ CT scanned, which revealed a defect porosity of 0.761% within the matrix material, but pore sizes were all below 0.0004 mm^3 and the vast majority under half that size. These defects were not evenly distributed, where size and count were topological feature dependent. The vertical struts were much more defect free than the struts at 45° angles (relative to the build direction), and the spheres exhibited the most significant porosity, likely due to keyholing and excessive heat accumulation from unsupported downskin surfaces. An SEM analysis was also performed for microstructure dependent sensitivity to geometric features. That analysis revealed the different supporting conditions drove unique thermal histories which resulted in competing grain-nucleation and grain-growth mechanisms. As for the structural efficiency, it was determined the mid-sized spheres with a radius of 0.75 mm performed best.

The ability to control the modulus of a material with architected lattice structures is particularly useful to medical implants where customizing the strength to best match different kinds of bone reduces the negative effect of a phenomenon known as stress shielding. Stress shielding occurs if an implant is much stronger than the surrounding bone, where most of the load will be taken up by the implant, and reduces the mechanotransduction of the bone. This loss of stress on the bone will result in bone loss or resorption, leading to the implant becoming loose, or failing. Therefore, the ability to tailor an implant modulus to cancellous bone or cortical bone,

which have moduli of 0.1-5 GPa and 15-20 GPa respectively, would be an enormous benefit, and is one of the key interests into medical applications of additively manufactured lattice structures [66].

Zhang et. al., designed a Rhombic strut-type lattice unit cell to mimic trabecular bone structure, and was to be manufactured from Ti-6Al-4V via EB-PBF [67]. The micro-architecture of the lattice was tuned by scaling and modifying strut lengths (1.75-2.5 mm) and diameters (0.128-0.342 mm) to create five different structures of varying porosity (87-96%). However, the as-built porosities ended up being lower than as-designed (61-80%) because the actual strut sizes all measured 0.400 ± 0.010 mm, indicating the minimum manufacturable feature size based on the process parameters. These designs were tested in compression and displayed a typical bending dominated failure mode with a reduced modulus of 93-96% from that of the bulk titanium. The authors were able to achieve apparent elastic moduli between 0.39-0.62 GPa which was similar enough to natural bone such that the stress shielding effect would be reduced.

Understanding mechanical anisotropy of lattice structures is important to good design, particularly for orthopedic implants. Therefore, Huang et. al. [68], investigated orientation dependencies of compressive fracture behavior on EP-PBF Ti-6Al-4V lattices. They looked at the combined effects of loading direction, cell orientation, and manufacturing irregularities, using a simple strut type cubic unit cell with a length of 1.1 mm, and 0.5 mm strut diameter, giving the structure a relative density of 0.36. The unit cells were oriented in the [001], [011], and [111] directions relative to the build direction. Predictably, the [001] orientation which had continuous struts directly in line with the loading direction, exhibited much higher strength, whereas the [011] and [111] were nearly equivalent. In terms of failure behavior seen in Figure 2.10, it was determined that the surface irregularities of the [011] and [111] specimens, created micro-notches

that acted as crack propagation sites for a plane of struts to fracture at low strain values. This work provided another example of the mechanical behavior of small-scale lattices being heavily defect driven rather than material or topology driven.

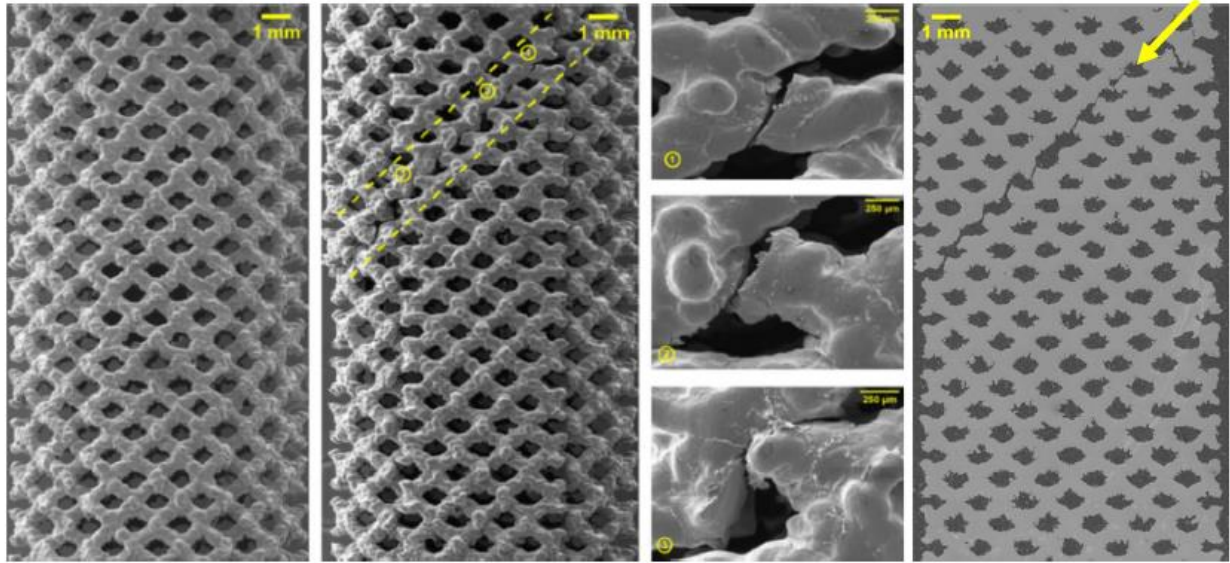


Figure 2.10. SEM images of a [011] orientation lattice untested sample (left) and a fractured sample (right), with closeups of three fracture locations and a cross section at the fracture location [68].

Two unique strut type lattice structures, an imitation saddle surface (ISS) and imitation arch bridge telescopic (IABT), were compared with a common Body-Centered Cubic (BCC), in an effort by Zhang et. al. to reduce the elastic modulus of L-PBF tantalum for medical applications [66]. Their designed porosity was 80% with an average pore size of $850\ \mu\text{m}$, and sized the strut diameter d and length l based on combinations of the Gibson-Ashby model, Euler-Bernoulli beam theory, Timoshenko beam theory, and Maxwell criterion. Relations were derived for the mechanical properties of elastic modulus, and yield stress shown in Equations (2.20) and (2.21)

$$\begin{cases} \frac{E^*}{E_s} = f\left(\frac{d}{l}, \theta, s\right) \\ E^* \propto \frac{d}{l}; E^* \propto \frac{1}{\theta}; E^* \propto s \end{cases} \quad (2.20)$$

$$\begin{cases} \frac{\sigma^*}{\sigma_s} = f\left(E^*, \frac{d}{l}, s\right) \\ \sigma^* \propto E^*; \sigma^* \propto \frac{d}{l}; \sigma^* \propto s \end{cases} \quad (2.21)$$

where E^* and σ^* are the elastic modulus and yield strength of the lattice, θ is the inclination angle of the strut, and s is the strut number.

These equations were used to design the cell geometries for a high yield-stress-to-elastic-modulus ratios. The average design length of the ISS struts was 0.69 mm compared with the BCC at 1.56 mm, though the diameters were similar at 0.34 mm and 0.39 mm respectively. The lattice structures produced for compression experiments were consistent with specimen requirements laid out in ISO 13314: 2011. Results of the experiments were compared with that of FEA, where the authors were successful at obtaining elastic moduli in the range of cancellous bone. The ISS ended up with the highest yield-stress-to-elastic-modulus ratio, double that of the traditional BCC, with a modulus of 0.63 GPa. Superior compression performance of the ISS configuration was attributed to the struts having a higher angle of inclination, a greater number of struts and nodes, and a larger diameter-to-length ratio.

Triply Periodic Minimal Surfaces (TPMS) get their name by being infinitely periodic in three dimensions (x, y, and z), and being a minimal surface that has a mean curvature equal to 0 (i.e. an implicit function $\varphi = 0$) which provides the highest surface area. As seen in Figure 2.11, these structures have smooth boundless surfaces that divide a volume into two labyrinths in the absence of self-intersections [69]. There are many benefits of TPMS architectures in that being mathematically derived present efficient methods for computing design parameters (size, shape,

volume fraction, channel interconnectivity). They also have smooth transitions at connection points to reduce stress concentrations and provide geometric continuity for better mechanical properties. For layer-based PBF additive manufacturing, the TPMS topology minimizes the staircase effect and nodal points seen in strut-based designs that result in greater fatigue resistance. In many applications like heat transfer, high surface-to-volume ratios are desirable in addition to separating the space into two interpenetrating sub-volumes. For multi-functional design of materials and systems, TPMS lattices appear to hold great promise in mechanical, thermal, electrical, acoustic, and photonic properties [49].

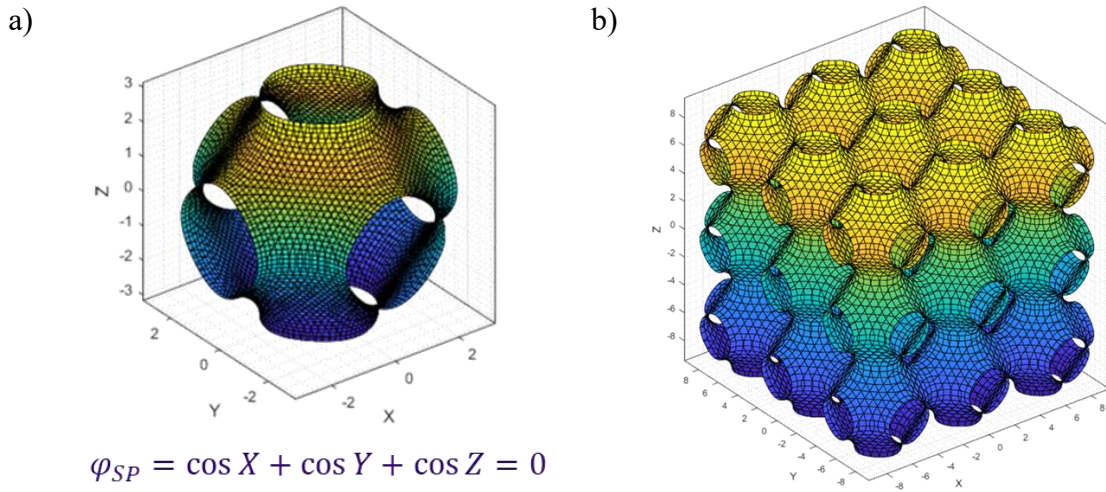


Figure 2.11. Example of the Schwarz Primitive TPMS (a) unit cell and (b) 3^3 unit cells plotted in MATLAB.

Shen et. al. [70], produced Ti-6Al-4V lattice structures resembling a Schwarz Primitive TPMS via SLM and investigated the potential influence of various strengthening structures during dynamic compression loading. The structures amounted to localized thickness increases at the unit cell level and therefore the authors were able to manipulate the structure failure location from the vertical struts to the horizontal joints. The reported 46.1% strength increase of the best

configuration did coincide with a roughly 9% increase in relative density over the basic structure, therefore the increase in specific collapse strength was not as significant.

In order to repair bone defects and mimic bone tissue, Liao et. al. [71] implemented a TPMS Gyroid scaffold with a radial gradient porosity distribution from L-PBF Ti-6Al-4V. Their CAD designed thickness at 150 μm was smaller than as-built thickness, which decreased the expected porosity from their three test samples that ranged from 333 μm to 674 μm , or 55% to 70% (porosity being the inverse of relative density). They reported from compression tests that the Young's modulus and yield strength ranged from 2.7-7.4 GPa and 233-520 MPa respectively.

A TPMS Gyroid lattice structure was investigated by Ma et. al. for how various design parameters affect biocompatibility and bone-repair in medical applications [72]. Five designs, with relative densities ranging from 11-25%, were produced from 316 L stainless steel by the L-PBF process. As shown in Figure 2.12, these were all very small as the unit cell size was varied from 1 mm to 2.6 mm, with a constant wall thickness of 100 μm ; however, the as-built samples had even smaller cells that measured 0.84 mm to 2.5 mm, and wall thicknesses that measured between 146 μm and 180 μm . At that scale, manufacturing defects drove experimental mechanical strength in compression to be 30-56% lower than FEA simulations predicted, highlighting the need for further research into the mechanical behavior of these lattice geometries. Permeability and cell growth experimental results demonstrated suitable biocompatibility, though larger porosity to obtain higher permeability would be preferable to promote bone repair.

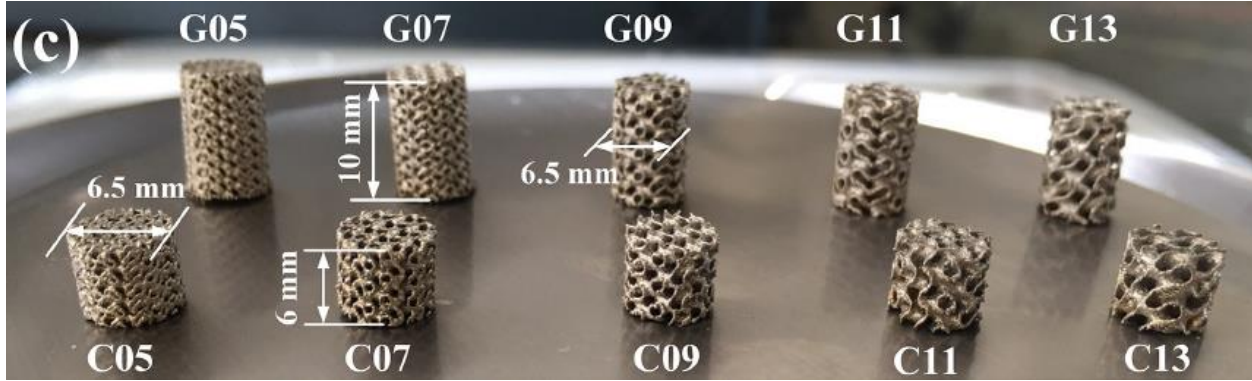


Figure 2.12. TPMS Gyroid samples with relative densities 11-25%, produced from 316 L stainless steel via L-PBF [72].

The authors Yan et. al. produced two studies in particular that investigated the microstructure and mechanical properties of an AlSi10Mg TPMS Gyroid lattice [73], and Ti-6Al-4V TPMS Gyroid and Diamond lattice structures for bone implants [74]. These lattices used the skeletal form of TPMS structures, as opposed to the sheet form, and employed volume fractions ranging from 5% to 20 %, with unit cell sizes that ranged from 3 mm to 7 mm. In the AlSi10Mg Gyroid they employed the Gibson-Ashby model in Equation (2.22) for compressive strength versus relative density to determine C_1 had a value of 0.8596 for all different cell sizes.

$$\frac{\sigma}{\sigma_0} = C_1 \left(\frac{\rho}{\rho_0} \right)^{1.5} \quad (2.22)$$

For Ti-6Al-4V, they again used the Gibson Ashby relations for determining modulus and strength versus relative density in Equations (2.23) and (2.24), and compared the Gyroid and Diamond structures with the model.

$$\frac{E}{E_0} = \left(\frac{\rho}{\rho_0} \right)^2 \quad (2.23)$$

$$\frac{\sigma}{\sigma_0} = 0.3 \left(\frac{\rho}{\rho_0} \right)^{1.5} \quad (2.24)$$

On log-log plots, they found good agreement with the analytical model such that they stated it is possible to design and precisely predict the mechanical properties of the aluminum or titanium TPMS lattices based on the volume fraction. They concluded the titanium study reporting that the Diamond and Gyroid TPMS lattices of 5-20% relative densities exhibited a modulus of 0.12-1.25 GPa and could be adjusted to match trabecular or cortical bone in order to avoid stress shielding.

The vast majority of mechanical behavior characterization for metal lattices has been conducted in compression where the structure deforms by a combination of bending, twisting, and stretching. In static compression tests the peak stress or Young's Modulus are typically plotted against the lattice relative density, as shown in Figure 2.13. Fatigue compression behavior is primarily determined by the interaction of strut cyclic ratcheting ($\partial\epsilon/\partial N$) such as the accumulation of strain from strut bending [75]. Fatigue crack initiation and growth through the lattice is driven by the surface roughness, but cell topology also plays a major role. It was reported that cubic lattices tend to have better fatigue strength than Dodecahedron lattices by a factor of seven. In some cases, three-point bend specimens were tested in metal with an open cell uniform lattice [76], or in polymer with a topology optimized graded lattice [69].

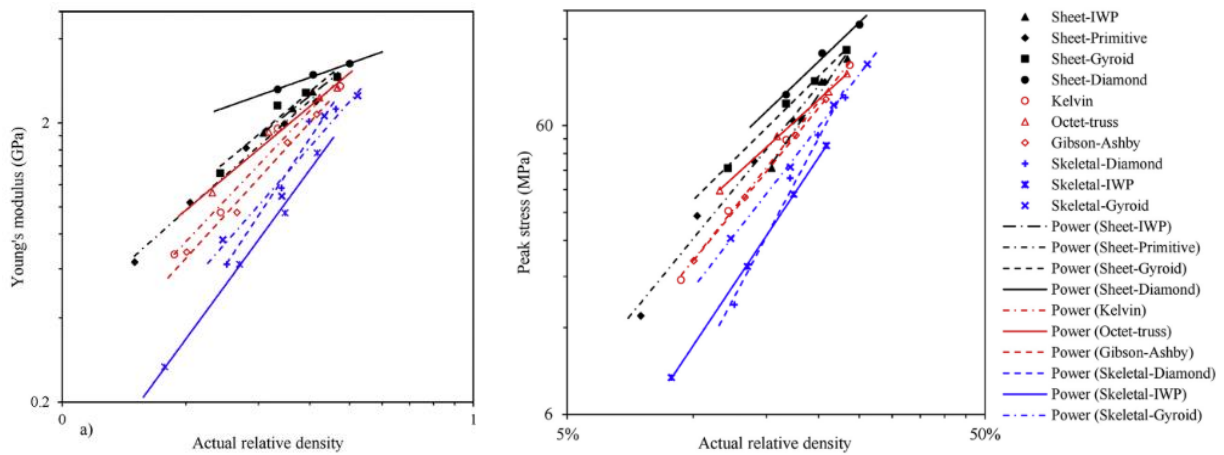


Figure 2.13. Experimentally derived Youngs modulus and peak stress versus relative density of various lattice geometries [56].

The effect of HIP treatment and strain-rate in compression on Ti-6Al-4V EB-PBF produced BCC and BCCZ microlattice structures was investigated by Ramadneh et. al [77]. Their specimens were designed to be 20 mm^3 , with 4 unit cells 5 mm^3 in each direction, and circular struts with a diameter of 0.5 mm . This resulted in the BCC and BCCZ microlattices having a relative density of $\rho_r = 0.075$, and $\rho_r = 0.087$, respectively. The specimens were subjected to quasistatic compression at low strain rates of 0.0005 s^{-1} and 0.05 s^{-1} , as well as drop impact tests to apply high strain rates of 75 s^{-1} and 150 s^{-1} . While printing such small lattices results in fast solidification rates and a very fine microstructure, they observed the HIP procedure had a coarsening effect that resulted in significantly higher Young's modulus, but only slight increases in compressive strength and energy absorption. The BCCZ, with its vertical struts, had the advantage over compressive strength, which increased over the BCC as strain rate increased; meanwhile, the BCC compressive strength exhibited a very low sensitivity to strain-rate. Figure 2.14 demonstrates how well their results compared with other strut lattices as a function of relative density.

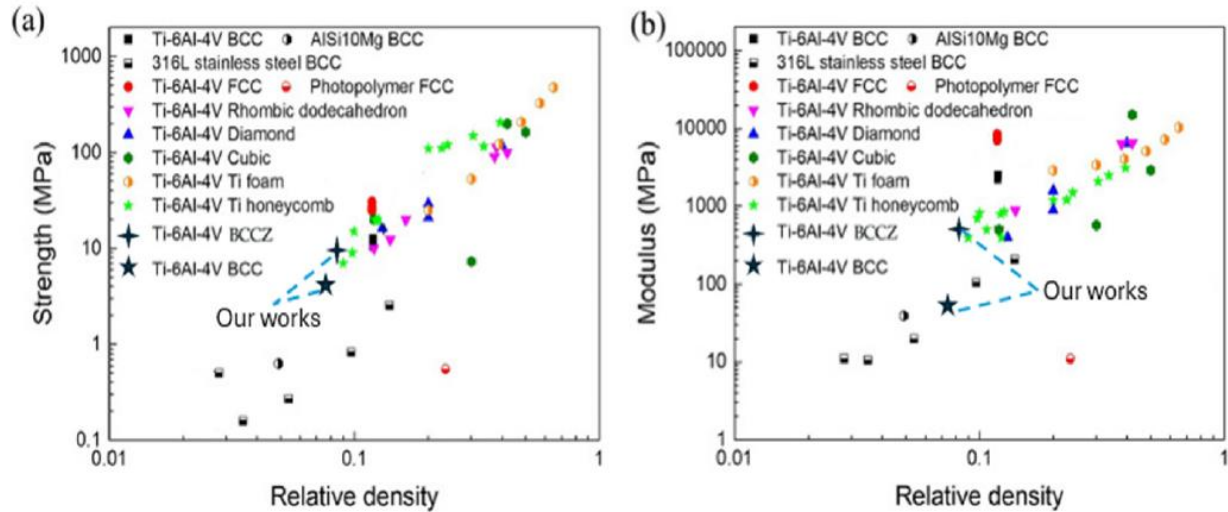


Figure 2.14. Mechanical properties of various cellular materials in a) compressive strength, and b) Young's modulus, as a function of relative density [77].

Residual stress in small models with fine geometry can lead to detrimental effects such as warping and cracking, thus Bruggeman et. al. investigated residual stress in Inconel 718 L-PBF lattice geometries [78]. The software platform Simufact Additive, which employs the eigenstrain reconstitution method (ERM) (i.e. inherent strain method), was used to simulate the building of plates and cubes containing TPMS lattice geometries to validate residual stress and deformation predictions. The TPMS unit cells investigated were the I-WP, Primitive, Diamond, Gyroid, and Lidinoid, and were integrated into plates of 25.4 mm x 25.4 mm x 1.6 mm and cubes of 25.4 mm, though the relative density and unit cell size of these was not reported. From model images, the cubes appeared to be an array 8 unit cells in each direction, meaning the geometries simulated were very small.

Nevertheless, the author's results showed that the lattice geometries had a significant impact on the residual stress distribution throughout the volumes with which they were placed [78]. It appeared that the Gyroid displayed the highest equivalent stress, while the Primitive exhibited the lowest. Simulated stress values reported during the L-PBF process were around 500 MPa, nearly 2/3 of YTS (758 MPa) for Inconel 718; however, comparison to other experimental studies showed the predicted values were much higher. This discrepancy is due to the flow stress curves for Inconel 718 employed by the software was meant for prismatic structures, and do not translate to complex lattice geometries.

The deformation modes of lattices under compression loading are identified as either stretching or bending dominated as shown in the Figure 2.15 diagram. Stiffness and strength tend to scale linearly with relative density, and they are higher for the stretching dominated than the bending dominated lattices that scale quadratically [56]. Stretching dominated deformation prevails if parts of the lattice structure carry loads via tension or compression and the exponent of

the stiffness-relative density curve is ~ 1 . Bending dominated deformation occurs if parts of the structure support loading by bending, twisting, or buckling, and the exponent of the stiffness-relative density curve is ~ 2 . In compression, the stretching dominated lattices usually exhibit a post-yield softening followed by a plateau stress, whereas bending-dominated lattices generally display a plateau stress right after yield [69]. It was found that sheet-TPMS topologies typically follow the stretching dominated behavior while strut topologies follow the bending dominated. In the case of the sheet-Diamond TPMS lattice architecture, a low exponent value of ~ 0.5 was reported which indicated a behavior that is nearly independent of relative density [56].

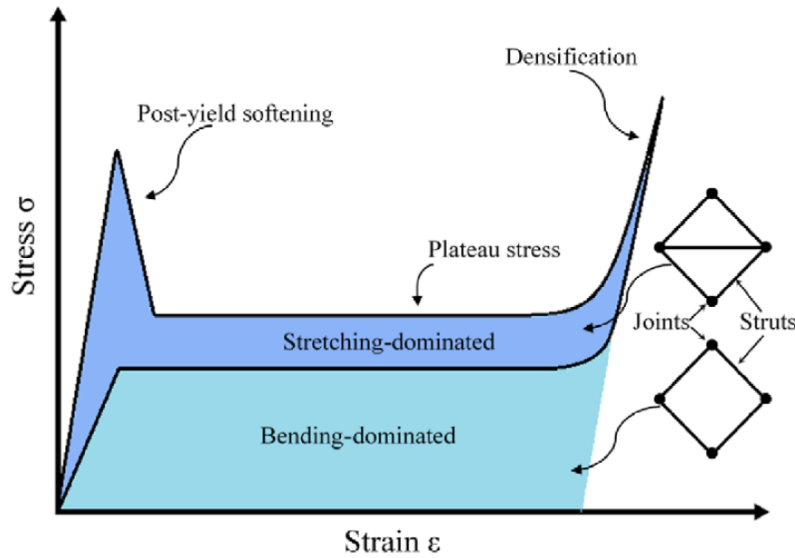


Figure 2.15. Diagram of the typical stress-strain compression behavior for bending and stretching dominated lattices [69].

Unit cell struts are often assumed to behave like slim straight Euler-Bernoulli beams, with equation constants determined by bending or stretching dominated behavior of the unit cell. Maxwell stability criterion M shown in Equation (2.25) can also determine the behavior for strut-based 3D lattices,

$$M = b - 3j + 6 \quad (2.25)$$

where b is the number of struts and j is the number of nodes [79]. If in the case of $M > 0$, then the structure is defined as stretch-dominated, and vice versa if $M < 0$, then it is said to be bending-dominated. While stretching-dominated lattices typically have a higher initial stiffness and yield strength than bending-dominated, they are prone to sudden failure where bending-dominated are more compliant. Because stretching-dominated struts are almost exclusively loaded in tension or compression, they are generally more structurally efficient than bending-dominated structures [17].

Sokollu, Gulcan, and Konukseven [80], compared the mechanical properties of EB-PBF Ti-6Al-4V strut and TPMS type lattice structures. They looked at compressive and tensile loading conditions of the BCC and FCC strut lattices (bending-dominated), and the Gyroid, Primitive, and Diamond TPMS lattices (stretch-dominated), to calculate the Gibson-Ashby relations and related coefficients according to Equations (2.26), (2.27), (2.28), and (2.29).

Deformation mechanism	Definition	
Bending-dominated	$\frac{E}{E_0} = C_1 \left(\frac{\rho}{\rho_0} \right)^2$	(2.26)
	$\frac{\sigma}{\sigma_0} = C_5 \left(\frac{\rho}{\rho_0} \right)^{3/2}$	(2.27)
Stretch-dominated	$\frac{E}{E_0} = C_1 \left(\frac{\rho}{\rho_0} \right)$	(2.28)
	$\frac{\sigma}{\sigma_0} = C_5 \left(\frac{\rho}{\rho_0} \right)$	(2.29)

This work used the nTopology platform to model the specimens per ASTM E8 or ASTM F3302 recommendations for tensile or compression experiments, respectively, and they correctly

identified 0.6 mm as the minimum producible thickness for EB-PBF. The target relative density was 0.25, but measured densities ranged from 0.256 and 0.282 for the BCC and FCC, and then 0.326, 0.340, and 0.385 for the Primitive, Diamond, and Gyroid.

In the tensile response [80], the TPMS types outperformed the strut types, with the Diamond exhibiting the highest strength, followed by the Gyroid, Primitive, FCC, and BCC, yet the Diamond also had the lowest strain to failure. That same trend was demonstrated in compression, though the FCC and BCC were very similar. The Gibson-Ashby C_1 coefficient for the FCC and BCC came out to be 0.46 and 0.56, while the Primitive, Gyroid, and Diamond were 0.14, 0.14, and 0.17. For the C_5 coefficient, FCC and BCC were reported as 1.30 and 1.17, with the Primitive, Gyroid, and Diamond calculated as 0.98, 1.01, and 1.23. Where the values of C_1 fell within the expected range of 0.1-4.0, C_5 values came in at or exceeding the range of 0.1-1.0, though no explanation for this was offered.

In a study conducted by Zhou et al. [81], the authors proposed a design strategy for self-supporting strut lattices, which considered an AM overhang constraint, on the basis of solid physics multi-fold symmetry of crystallography. They designed lattice unit cells with four hierarchical levels composed of angled and vertical struts, and applied periodic boundary conditions. Then using the elastic deformation-constitutive relation, they derived the unit cell stiffness matrix $\mathbf{K}_{uc}$ shown in Equation (2.30).

$$\mathbf{K}_{uc} = \sum_{i=1}^n \mathbf{T} \tilde{\mathbf{K}}^{e(i)} \mathbf{T}^T \quad (2.30)$$

By evaluating the reversibility of the stiffness matrix, such that if Equation (2.30) is not reversible, then the lattice unit cell would be bending-dominated; and thus the opposite, if it is reversible, then the lattice is stretch-dominated. From this criterion, the authors determined analytically that their

first and second hierarchical level unit cells were bending-dominated, while their third and fourth hierarchical level unit cells were stretch-dominated.

A data review of available mechanical properties for metal PBF lattice structures was published by Hanks et. al. [82]. They attempted to create a database called the Lattice Unit-cell Characterization Interface for Engineering (LUCIE) to allow researchers to compare mechanical properties of different unit cell topologies. Examples of the more common ones were the Body-Centered Cubic (BCC), Face-Centered Cubic (FCC), the Octet truss, Schoen Gyroid, and the Schwarz Diamond. However, the endeavor found information to be rather limited as their initial search of 117 articles ended up being down selected to just 69. While the information was categorized by unit cell topology, differences in processing conditions and material created a wide range of potential property values, and it was noted that many researchers did not follow standardized guidelines for test samples. Additionally, there are an infinite number of minor variations that can be performed on a unit cell topology, such as rounding and filleting, that make it impossible to identify and track each one. Therefore, this study revealed the large gap that still remains in determining the potential for AM lattice structures, and the importance of focusing on testing quality material by documented standard methods.

Jonnala et. al. [83] identified a gap in the exploration of unit cell design and investigated novel lattice structures to provide optimal energy absorption using the polymer fused filament fabrication (FFF) process. They used a MATLAB to generate three unique TPMS-like isosurface based on user input criteria, and employed nTopology to render them as 3D solids. The unique topologies were compared against the TPMS Gyroid and TPMS Diamond, shown in Figure 2.16, all with relative densities around 50%. Under compression loading the Diamond topology proved equal in yield strength compared to the best optimized lattice, though in terms of energy absorption, the

three designed specifically for that purpose showed superior. This work highlights the opportunity space for customized lattice topologies to demonstrate higher performance.

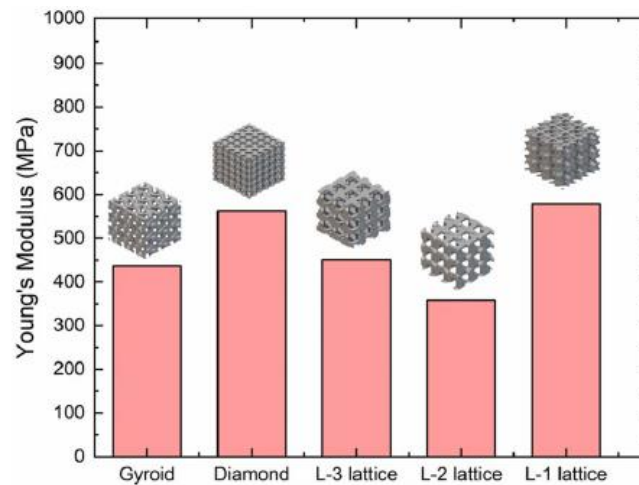


Figure 2.16. Young's Modulus for three unique TPMS-like isosurface L-1, L-2, L-3, with Gyroid, and Diamond structures, relative densities of 50%, fabricated via polymer FFF [83].

Di Frisco et. al. leveraged TPMS unit cells to explore new topological designs that control their mechanical properties [84]. They used the TPMS-sheet Primitive to develop a hierarchical hybrid unit cell with an integrated smaller-scale skeletal-TPMS Gyroid, IWP, or Primitive structure. The implicit modeling platform nTopology was employed to create the geometry and mesh the structures, while the predictive numerical analysis was conducted with Abaqus, and compared with compression experiments. Cubed specimens of 3-unit cells in each direction, and measured 45 mm^3 , were produced via FFF from PLA. They found the original Primitive cell possessed anisotropic behavior, with lower stiffness in axial directions. Thus, the second order lattice functioned as reinforcement, and provided a more even stress distribution that increased yield stress and elastic modulus. The hybrid structure reached the plateau regime at a higher stress; although, this also led to an earlier onset of densification such that it did not always result in an increase of total energy absorption.

Xin et. al. [85], published a 2025 literature review of fatigue failure mechanisms and life prediction of metal AM lattices. One point of interest identified was that the skeletal-TPMS Diamond lattice had a longer lifespan compared to the Gyroid because the tensile-dominated Diamond structure had fewer bending components, and slower ratcheting rates, than the bending-dominated Gyroid. It was also noted that sheet-TPMS typically demonstrated higher fatigue strength compared to the skeletal, and the skeletal-TPMS typically outperforms strut-based like the BCC. Of strut-type structures, those with struts aligned in the loading direction, composed primarily of optimal print quality, and designed to minimize localized stress concentrations, demonstrated superior fatigue performance. The number of struts determines the loading type, and in bending-dominated lattices the tensile stresses generated at the strut surfaces lead to crack initiation. With tensile-dominated lattices, the strain accumulation rate is generally lower with the same relative density, which results in better fatigue performance. However, one of the biggest challenges identified with comparing study results is the lack of commonality and standardization between experimental specimens and testing methods.

Throughout the reviewed literature, consistent and reliable experimental data seemed to be difficult to produce, and important information is still missing for lattices, especially newer TPMS topologies. In terms of EB-PBF Ti-6Al-4V, reporting of stretch dominated TPMS lattices, that offer higher modulus and ultimate compression stress (UCS) at similar densities, is insufficient and the applicability of the Gibson-Ashby model is not well understood [47]. Many studies have focused on the smallest manufacturable geometries for medical purposes (450-1000 μm), yet structures at that scale are very susceptible to manufacturing defects. This implies that the overall fabrication of lattices via EB-PBF lacks understanding, where little attention has been given to producing quality structures that would translate to better material and mechanical properties. The

effect of process parameters plays a critical role in manufacturing quality, so it was surprising that has not been the focus of a large number of studies involving lattice architectures.

2.2.2 *Graded Lattices*

While different lattice topologies have brought about a significant increase in geometric complexity, the complexity can be further expanded through functional grading. Unlike a uniform, or homogeneous lattice structure, a graded lattice varies the unit cell size and shape to change the relative density across the domain, and can be adapted to a particular function. To an additive PBF machine, the only impact of this additional complexity is potentially powder removal if that aspect is neglected during the design phase. Therefore, a uniform lattice does not make much sense to industrial applications when a weight optimized component, with a tuned relative density, would perform better at no additional fabrication cost.

Because functionally graded lattice structures have been identified as a way to regulate mechanical properties, they have therefore have been investigated for orthopedic implants. Zhang et. al. employed triply periodic minimal surfaces (TPMS) and heterogeneous functionally graded scaffolds (FGS) in order to blend skeletal and sheet TPMS-based scaffold with increasing density for a biomechanical gradient similar to trabecular or cortical bone [86]. Three of the more common TPMS Equations (2.31), (2.32), (2.33) and (2.34), are shown here for Diamond, Gyroid, I-WP, and Primitive configurations,

$$\phi_{\text{diamond}}(x, y, z) = \cos(ax) \cos(ay) \cos(az) - \sin(ax) \sin(ay) \sin(az) \quad (2.31)$$

$$\phi_{\text{gyroid}}(x, y, z) = \cos(ax) \sin(ay) + \cos(ay) \sin(az) + \cos(az) \sin(ax) \quad (2.32)$$

$$\phi_{\text{I-WP}}(x, y, z) = \cos(ax) \sin(ay) + \cos(ay) \sin(az) + \cos(az) \sin(ax) \quad (2.33)$$

$$\phi_{\text{primitive}}(x, y, z) = \cos(ax) + \cos(ay) + \cos(az) \quad (2.34)$$

where the constant coefficient a affects the period of the TPMS structures, such that a value of 1.33π would generate a unit cell size of 1.5 mm. Equation (2.35) identifies the solid region of a TPMS structure through the manipulation of the parameter t .

$$\phi_{\text{TPMS}}(x, y, z) + t \geq 0 \quad (2.35)$$

To create a smooth S-shaped transition gradient $G(x, y, z)$, a sigmoid function was used to express a graded skeletal TPMS-based FGS as in Equation (2.36), with the sigmoid function $\alpha(x, y, z)$ in Equation (2.37) to achieve a relative density gradient with the control parameter k .

$$\phi_{\text{FGS}} = \alpha(x, y, z)(\phi_{\text{TPMS}}(x, y, z) - t_0) + (1 - \alpha(x, y, z))(\phi_{\text{TPMS}}(x, y, z) - t_1) \quad (2.36)$$

$$\alpha(x, y, z) = \frac{1}{1 + e^{-kG(x, y, z)}} \quad (2.37)$$

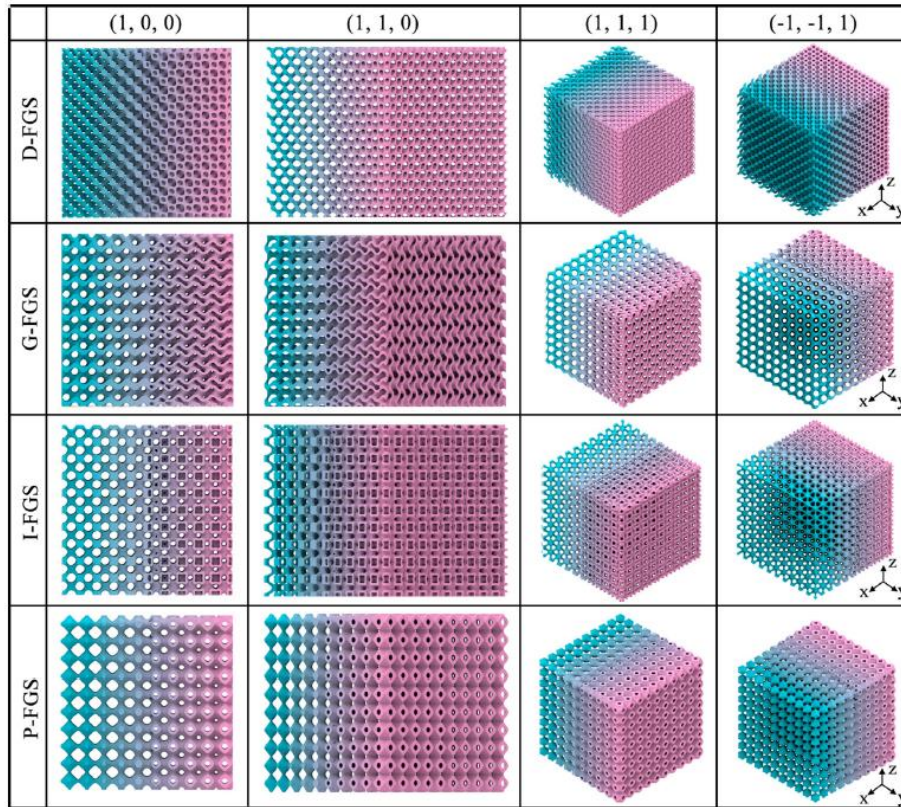


Figure 2.17. Graded Diamond, Gyroid, I-WP, and Primitive TPMS lattices in various directions [86].

With this method [86], the authors were able to design graded TPMS structures that transitioned smoothly from one geometry to another with variable density, as shown in Figure 2.17. They produced 15 mm³ samples with relative densities ranging from 25-55% via L-PBF from Ti-6Al-4V powder, and subjected the samples to a subsequent heat treat at 900°C for 2 hours to affect the titanium alloy microstructure. The thickness of the structures ranged from 0.216 mm to 0.457 mm, which are on the edge of the producible limits for this manufacturing process. It was suspected that defects were the cause for some reported premature failures; however, under small compressive strains any defects did not significantly alter the linear elastic response.

Per the analytical model presented by Zhao et. al. [51], and displayed in Equations (2.38) and (2.39), Zhang et. al. found it to produce accurate results for the apparent elastic modulus E_a and strength σ_a , with f_i , E_i , and σ_i representing the composing lattice structure volume fraction, elastic modulus, and strength, respectively [86]. The terms f_x , E_x , and σ_x represent the volume fraction, elastic modulus, and strength at a distance x along the length of the structure.

$$E_a = \sum_{i=1}^n E_i f_i = E_x \cdot f_x \cdot dx \quad (2.38)$$

$$\sigma_a = \sum_{i=1}^n \sigma_i f_i = \sigma_x \cdot f_x \cdot dx \quad (2.39)$$

Mechanical performance was evaluated by compression experiments according to ISO 13314, which found that the ultimate strength of sheet TPMS lattices generally outperformed the skeletal types, particularly the Gyroid possessed a more stable response and higher strength behavior. Overall, coupling skeletal and sheet TPMS lattice structures with functionally graded scaffolds demonstrated good potential to tailor the mechanical behavior locally and globally for superior performance.

Yang et. al., identified in their 2019 article [87], that existing data on the structural response of TPMS Gyroid lattice structures was sparse, which led them to perform numerical studies of the mechanical properties manufactured from L-PBF Ti-6Al-4V. They used the Taguchi, and One Factor at a Time (OFAT) statistical methods, to explore the effects of different geometric parameters, such as surface thickness, bulk size, number of surface periods, and associated isovalue. Then, applied their results to theoretically develop a Gyroid lattice structure that mimicked human cortical bone elastic modulus and compressive strength. The cell sizes and wall thicknesses were varied from 8 mm to 20 mm, and from 0.3 mm to 0.6 mm, respectively, filling a constant cube volume with a length of 40 mm. The different isovalues that were applied in this study were 0 (standard), 0.5, 0.7, and 1.1, where isovalue results in significant variation in the TPMS geometry as it is the value the Gyroid function is set equal.

The authors determined the ranking order for the geometric parameters with the largest influence on stiffness and strength to be: 1) number of cells, 2) surface thickness, 4) bulk sizes, and finally 4) isovalue [87]. Changing the isovalue from 0 ended up decreasing the strength and significantly altering the stress-strain behavior, while increasing the surface thickness resulted in increased modulus and compressive strength. They found that balancing cell size and thickness provided the opportunity to manipulate the Gyroid lattice modulus, maximum stress, and energy absorption. To that end, it was predicted through simulation that 7 cells per 40 mm bulk cube length, a wall thickness of 0.55 mm, and an isovalue of 0, would provide a yield stress of 150.47 MPa and an ultimate compressive strength of 300 MPa, meeting the criteria of human cortical bone.

Mechanical properties of homogeneous and gradient TPMS-skeletal Gyroid, and Diamond lattice structures, were investigated by Liu et. al. to match the elastic modulus of the human tibia

[88]. Relative densities of 50%, 60%, and 70% were used for the homogeneous designs, while the gradient design used a linear function that varied the relative density from 50%-70% in the +Z-direction, which resulted in an average of 60% density. These 15 mm cubed designs were simulated by FEA, and produced from Ti-6Al-4V via L-PBF for compression experiments. The measured modulus of the designed structures met the requirements for orthopedic implants, which ranged from 6.24 to 10.5 GPa. The linear gradient structures demonstrated higher modulus (66.9%), yield strength (20.3%), and compressive strength (16.3%), to the homogenous structures of comparable density.

Of particular note was that the authors claimed the Gyroid exhibited superior nominal stiffness compared to the Diamond [88]; however, this is based on a FEA strain range of 0.35-0.4, well after plastic deformation had ensued, and beyond the initial softening. The reported yield stresses were based on calculations at 0.2% plastic strain, yet that is criteria typically used for tensile measurements, and the FEA stress-strain plots actually show yielding occurring at or below 0.05%. In terms of modulus, which is typically used for comparing stiffness, the Diamond in all design cases was predicted to be 10-15% higher than the Gyroid. However, only the Gyroid specimens were used in compression experiments, so no empirical validation of their claim was provided. Furthermore, it was left unclear as to why the higher density structures failed at lower compressive strengths than the 50% dense, and that the actual modulus of the different densities were the same, yet 3-4 times lower than the FEA predicted.

The compression behavior of functionally graded Gyroid lattices made from 17-4 PH steel was studied by Caiazzo et. al. [89]. Using the implicit modeling software nTopology, they designed Gyroids with 15%, 25%, and 35% relative densities, with three different grading methods implemented in the build direction called uniform, thickness graduation, and void size graduation.

Their specimens were manufactured via L-PBF with wall thicknesses that ranged from 0.5-0.9 mm and unit cell sizes of 5.5 mm for the 35% dense, to 13 mm for the 15% dense. At these sizes, the experimental performance results saw the thinnest regions collapse first where surface defects have significant influence. A numerical analysis performed with COMSOL showed similar behavior due to the high stress concentrations in the thinnest regions, though it did not account for the effect of defects. With these grading methods it is not surprising that the uniform lattices performed best, but their results allowed for the computation of Gibson-Ashby parameters for C_1 , n , C_2 , m , and α , and provided a model to predict the bending-dominated behavior of samples graded in these ways.

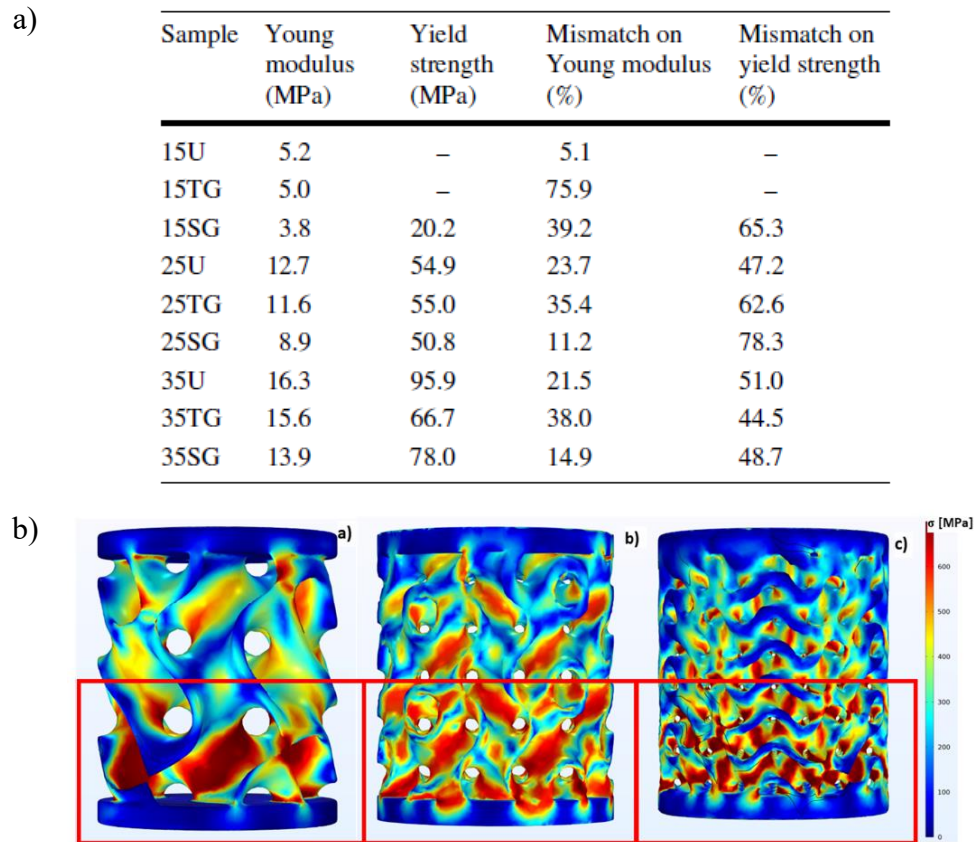


Figure 2.18. Graded TPMS Gyroid a) properties based on simulation and percent mismatch to empirical values, and b) stress distributions for various thickness gradings [89].

The mechanical behavior of multidimensional gradient TPMS Gyroid structures under static and dynamic loading was reported by Lu et. al. [90]. They wanted to have the ability to mimic complex gradient pore characteristics, similar to bone, and proposed a novel method of functional grading based on topological constants rather than the more common thickness, cell size, and porosity properties. Specimens were fabricated via L-PBF and tested in compression at rates that varied from 5 m/s to 50 m/s. Results showed that for static compression, deformation and collapse would first occur in the lower density regions, but that at higher speeds inertia effects would dominate in the form of localized deformation. However, some of the graded configurations demonstrated the ability to deform much more uniformly across the broad range of speeds.

As interest in printed lattice structures has grown, researchers like Ekeh et. al. [91] have been looking into the flexural properties of functionally graded lattices and hybridized with different topologies. Their study involved printing porous lattice beams from L-PBF AlSi10Mg for four-point bend testing, with topologies that included TPMS sheet types, such as the Schwarz Primitive, Diamond, Schoen Gyroid, Schwarz Diamond, Neovius, Schoen I-Wrapped Package, Fischer Kock, and Schoen F-Rhombic Dodecahedral, alongside the Octet strut type for comparison. While they demonstrated various grading methods with relative densities varying from 5% to 40%, they combined that with morphing from one lattice to another in the case of their Primitive-Diamond-F-Rhombic Dodecahedral specimen. That study showed how they were able to control the mode of failure in bending-dominated structures with relative density grading and hybridization (multi/poly-topology). They observed how uniform TPMS lattice beams would fail in shear while multi-topology TPMS beams, and example of which is displayed in Figure 2.19, failed in a tensile mode near the mid-span. Where the hybridized lattice structures resulted in a higher specific flexural modulus, their specific flexural strength and toughness were lower than the single uniform

or graded lattices. Finally, the authors note the importance of using stretching-dominated lattice topologies with grading when employed as a lightweight beam structure.

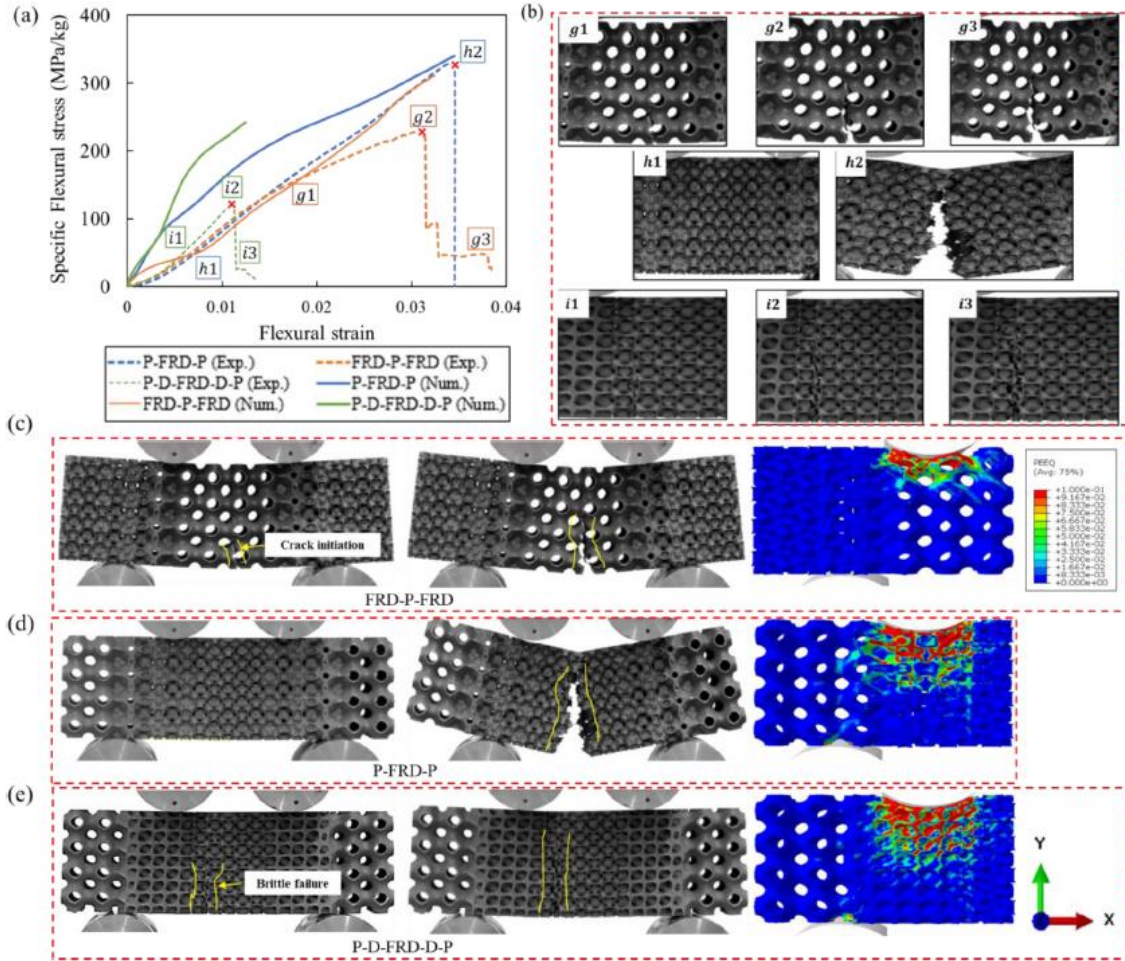


Figure 2.19. Example of multi-topology hybridized TPMS-sheet latticed-beams under 4-point bend demonstrating specific flexural stress-strain responses alongside simulations [91].

Abdelaal et. al. [92], also investigated the flexural behavior of lattice beams built from L-PBF Ti-6Al-4V. They experimented with combining density grading with conventional-auxetic lattice systems to create a density/functionally graded hybrid Poisson's ratio (FGHPR) lattice structure, and compared that with three other uniform lattice structures: a hexagonal unit cell with a positive Poisson's ratio (PPR), a re-entrant hexagonal unit cell with a negative Poisson's ratio (NPR), and

a hybrid (HPR) of the two unit cells combined at the lengthwise centerline of the beam. The thought behind the combination was that because both the hexagonal and re-entrant structures would undergo transverse contraction when subjected to a bending load, that would translate to significant beam thinning, hence a rise in densification, an increase in stiffness, and a subsequent higher plastic energy dissipation. The author's 4-point flexural experiments showed that the PPR lattice exhibited the highest flexural stiffness and peak load, followed by the NPR, HPR, and FGHPR which had about half the stiffness of the PPR. Yet, the FGHPR did have nearly twice the absorbed energy of the PPR, and the functional grading provided an improvement of the uniform HPR lattice.

2.2.3 *Topology Optimization*

The process of structural optimization, particularly in industries where weight is critical, often includes the use of topology optimization in early stages of concept development. Topology optimization is an iterative approach that seeks to uncover the best distribution of material within a specified volume. It is run as a finite element model with portions of the design domain designated as either design space (i.e. the volume to be optimized), or non-design space which are fixed features typically employed to impart loads or apply boundary conditions to the design space.

The most common topology optimization method was originally proposed by Bendsøe [93] and is called the SIMP method, which stands for "Solid Isotropic Material with Penalty." In this method, a multiplier on the element stiffness is designated as the design variable, and can change for each finite element in the design space region between zero and one. This design variable is commonly called the density, and the method is sometimes referred to as the density method, or pseudo-density method. The main philosophical idea behind SIMP, as seen in Equation (2.40), is

that the modulus of the material is reduced in a uniform way (hence “isotropic”) as the material becomes more porous:

$$E(\rho) = (\rho)^p E_0 \quad (2.40)$$

where the design variable ρ is the pseudo density (modulus/stiffness knockdown factor), $E(\rho)$ is the effective modulus of the element as a function of the pseudo-density, E_0 is the material modulus of elasticity, and p is the penalty factor ($p > 1$) intended to force the stiffness towards the extremes of fully solid or fully void.

In most engineering applications, the stiffest design with the least deflection under maximum loading is desired. To this end, the objective function for traditional topology optimization is to minimize the compliance, or global strain energy in the model, which is the area under the linear portion of the stress-strain curve. A constraint on the optimization is usually a user specified volume fraction that is an assumed target quantity of material to carry the loads. The problem the optimization is being asked to solve is, “calculate the stiffest arrangement possible using only ‘this much’ material.” A penalty factor is required because the optimization algorithm is looking to make elements either fully dense (pseudo-density = 1) or fully void (pseudo-density = 0) in a binary fashion, but a gradient based optimizer can only solve a continuous function that can be differentiated. Therefore, while the density varies smoothly between 0 and 1 it is exponentially “penalized” forcing the algorithm to push the design variable towards the end points of 0 or 1 with a minimal amount of difficult-to-interpret, partially dense material in-between [94].

Panesar et. al. offered up some strategies for using topology optimization to derive functionally graded lattice structures [95]. Based on a SIMP generated solution, they proposed: 1) intersect the solid solution with a uniform density lattice, 2) grade the lattice with a variable density map onto the pseudo-density solution with a defined upper and lower bound, and 3) scale a variable

lattice by mapping it to a re-scaled pseudo-density solution defined by an upper and lower bound. Those strategies were compared with both a solid topology optimized solution, and a homogeneous material that filled the design domain. They used the well-known simple 3D cantilever beam problem as the design domain, shown in Figure 2.20, and ran several solutions with volume fractions of 0.3, 0.5, and 0.85, where after being intersected with a lattice, resulted in an averaged structural volume fraction of 0.5. While this study did not perform any physical experiments to correspond with the numerical predictions, the analysis indicated the scaled method to provide a more robust structure than the graded or intersected strategies.

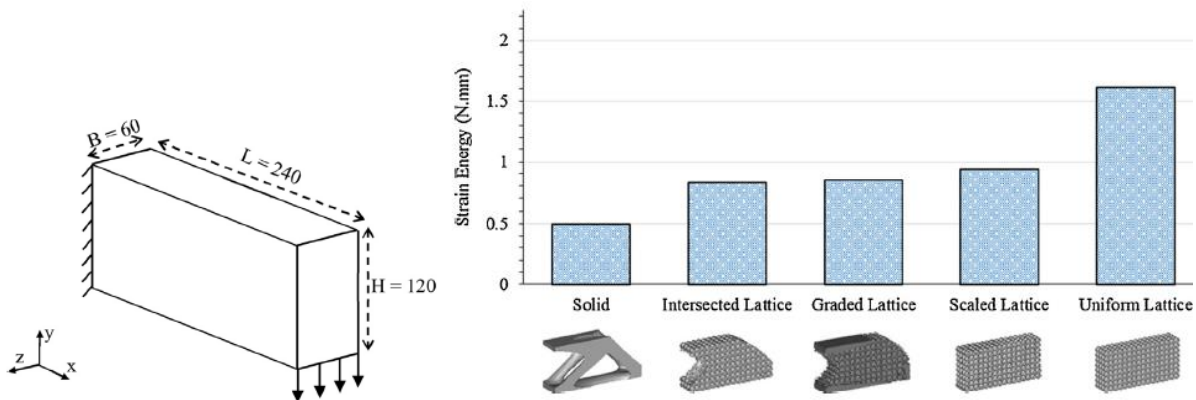


Figure 2.20. Cantilever beam problem and the total strain energy of 5 solutions using different lattice methods [95].

There are many more modern topology optimization methods that have been developed since the introduction of SIMP, but they are not as mature and have yet to see widespread implementation into commercial software packages. Some of these include heuristic-based evolutionary structural/topology optimization, the level set approach, and multi-component layout design, all with the goal of improving the refinement of the output over the traditional density-gradient method which is often very coarse and difficult to interpret to a CAD model. Increasingly, these other methods are being combined to take advantage of their different strengths and given

names such as Moving Morphable Voids, or Moving Morphable Components [94]. Besides static structural layouts, topology optimization algorithms are also being employed to solve dynamic, thermal, and manufacturing limitations, often employed as additional constraints on top of the volume fraction.

An aerospace case study of topology optimization for a helicopter was presented by Süß et. al. [96]. The component, which had to support the main gearbox, was optimized to save weight, and had to consolidate an assembly into a single part, while satisfying high-cycle and low-cycle fatigue requirements. This study highlighted the challenges and time-consuming nature of the topology optimization workflow, where the reconstruction of the optimized shape required detailed interpretation of the mesh surface back into a CAD model. Two different methods of shape conversion and refinement were executed, as displayed in Figure 2.21, one used a STEP model export and the other a STL model export. While the STEP model optimization method started with a 44% weight reduction, it gained weight due to the difficulty of the redesign process and ended with a lower 40% weight reduction. By comparison, the STL optimized shape provided a 58% weight reduction, which was able to be maintained following the subsequent redesign steps. To prove the designs were producible, they were fabricated via EB-PBF and showcased the potential for optimization of future aerospace structures.

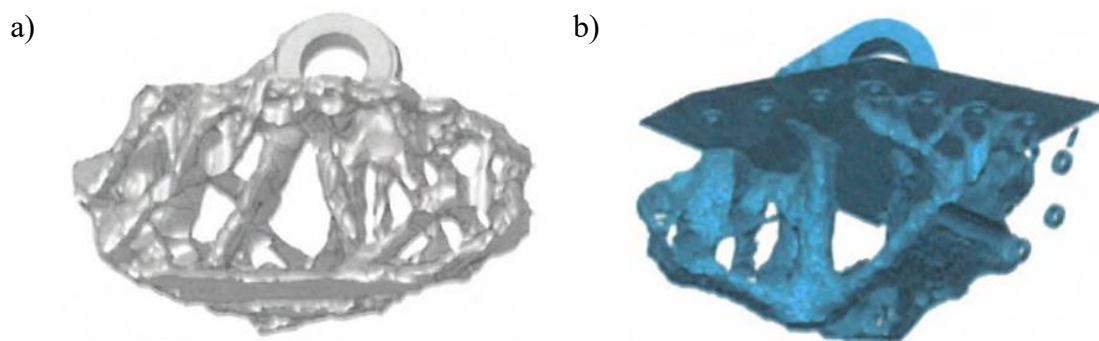


Figure 2.21. a) Topology optimization solutions for weight with output formats as a) STEP, or b) STL [96].

In the study published by Wu et. al [97], the authors implemented a multiscale gradient-based topology optimization method to design an injection mold tool with conformal cooling channels to be manufactured via L-PBF. The proposed method at the macroscale sought to minimize the mass of the mold, while at the mesoscale satisfied the thermomechanical constraints. What they described as a 3D porous injection mold, implemented strut type lattice unit cells that optimized bulk modulus and shear modulus to minimize compliance at the same time as maximizing heat conduction. In this application the mass fraction (i.e., relative density) was initially defined as 50% before it was optimized to the final structure of 71%.

Zhao and Zhang [98], optimized straight and curved beam structures with multiple levels of porosity using a modified SIMP-like scheme. In order to demonstrate structural robustness and fail-safe behavior they also included some simple local damage models to simulate a type of material defect. Essentially, this was done by removing material from certain locations defined by a circle sized to match the member size of the highest-level bulk design. By optimizing around these defects, and merging the solutions, their beams specimens fabricated out of photopolymer resin, demonstrated relatively high structural stiffness and load-carrying capacity in 3-point bend experiments. This work helped with identification of differences in the ductility of failure modes between uniform and topology optimized structures.

One of the challenges topology optimization brings to structural design is the continuously variable, organic-like, geometric beam cross sections that make traditional strength analysis difficult. The second area moment of inertia I in classical beam theory, formalized by Euler and Bernoulli in 1750, measures how well a beam's cross-section will resist bending based on its shape. However, as Huie et. al. described in their study [99], they developed a software tool called SegmentGeometry. It is an extension to the open-source 3D Slicer platform, designed to measure

the second area moment of inertia of biological structures based on CT scans, such as a salamander leg bone shown in Figure 2.22.

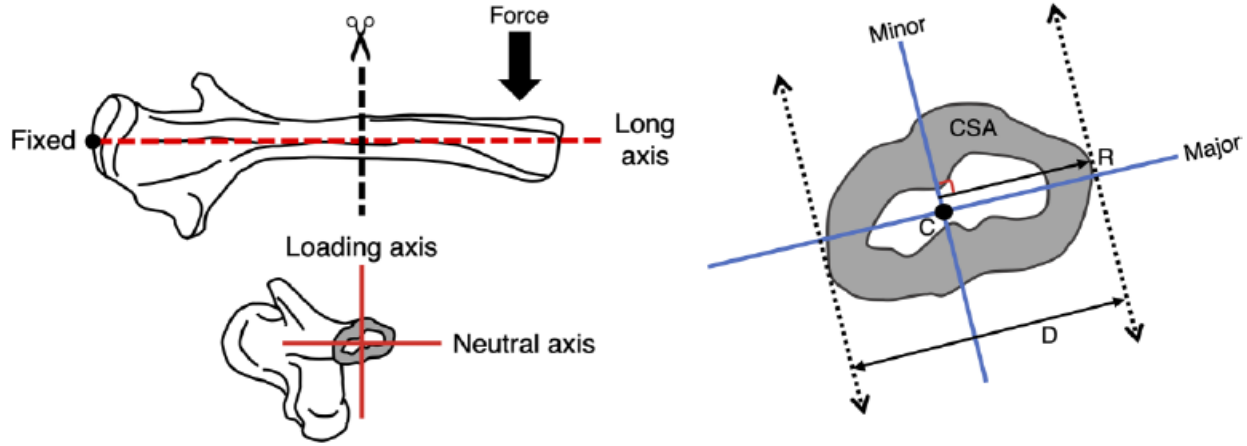


Figure 2.22. Illustration of a bone in cantilevered loading, with a cross-section at the mid-point to show the orientation of the loading and neutral axes, where measurements can be taken to include the cross-sectional area (CSA), centroid (C), distance to the furthest point from the minor axis (R), and maximum Feret diameter (D) [99].

The second area moment of inertia is calculated about the neutral axis according to Equation (2.41),

$$I = \int x^2 dA \quad (2.41)$$

where x is the distance between the neutral axis and an area dA . SegmentGeometry used Equation (2.42) for the calculation of I for an arbitrary composite shape around the neutral axis,

$$I = \sum_{k=1}^n \int I_k + A_k D_k \quad (2.42)$$

with A_k as the pixel area, D_k as the perpendicular distance between the neutral axis and the arbitrary axis, and n as the number of pixels in the cross-section. A common data format for CT scans are STL files, which is the same as those output by CAD software to prepare 3D printer build

files. In the same way this tool can determine the second area moment of inertia for a biologic structure, SegmentGeometry could also evaluate it for a continuously variable topology optimized lattice structure.

A phase-field approach to topology optimization was presented by Takezawa et. al. [100], who sought to optimize a lattice unit cell topology that would be isotropic in the unit vector directions of 100, 110, and 111. They also included a DfAM constraint with powder removal holes in the search parameters since specimens were to be manufactured via EB-PBF. The phase-field method uses an initial ideal state called the density function ϕ that defines $\phi = 1$ as solid, and $\phi = 0$ as void, for the unit cell. Then, after using FEM to determine displacements and the effective elastic tensor, the algorithm calculated the sensitivities relative to the objective function and the constraint shown in Equations (2.43) and (2.44) respectively.

$$\text{Minimize: } -\kappa^H + P_{iso} \quad (i = 1, 2, 3) \quad (2.43)$$

$$\text{Volume Constraint: } \frac{\int_{\Omega} \phi dx}{\int_{\Omega} dx} \leq V_{max}, \quad 0 \leq \phi \leq 1 \quad (2.44)$$

The objective function minimizes the difference between an effective stiffness determined by the bulk modulus κ , and the stiffness of the isotropic state P_{iso} , which is similar to a compliance minimization function. The solution space Ω was limited by a prescribed volume constraint V_{max} that is found in most topology optimization algorithms. The authors also used the Hashin-Shtrikman upper bound relation [101], shown in Equation (2.45), as a check on the solution. This is based on the Rule of Mixtures where κ is a function of volume fraction f so as to set a limit on the theoretical stiffness.

$$\kappa = f\kappa_{Ti} + (1 - f)\kappa_{Air} - \frac{f(1 - f)(\kappa_{Ti} - \kappa_{Air})^2}{(1 - f)\kappa_{Ti} + f\kappa_{Air} + \frac{4G_{Ti}}{3}} \quad (2.45)$$

Takezawa et. al. then employed the Allen-Cahn relation [102], i.e. a diffusion-reaction Equation (2.46), to evolve the density function based on the constraints. In this way, $\kappa \nabla^2 \phi$ served as the diffusion term that drives towards space homogenization, while $P'(\phi)$ provided the reaction term that could increase or decrease depending on the slope of Equation (2.47). The phase-field technique depends on the multi-well term $Q(\phi)$ in Equation (2.48) to stabilize the boundaries, and the interpolation function $R(\phi)$ in Equation (2.49) to penalize regions where $0 < \phi < 1$.

$$\frac{\partial \phi}{\partial t} = \kappa \nabla^2 \phi - P'(\phi) \quad (2.46)$$

$$P(\phi) = \frac{1}{4} Q(\phi) + \eta R(\phi) \quad (2.47)$$

$$Q(\phi) = \phi^2(1 - \phi^2) \quad (2.48)$$

$$R(\phi) = \phi^3(6\phi^2 - 15\phi + 10) \quad (2.49)$$

Through this unique approach, the study examined volume fractions of 20%, 25%, and 30%, and depending on the location of the powder removal holes, ended up with geometries resembling the Schwarz Primitive TPMS.

Wang and Sigmund [103] utilized a flexible framework for large scale topology optimization to extend 2D material design with enhanced buckling strength into 3D design with tunable stiffness and buckling response. Their work also took place at the unit cell level with periodic boundary conditions. They started with a hollow sphere at 20% volume fraction and sought to optimize the geometry in hydrostatic and uniaxial compression. The unit cell design space was discretized by 64^3 11-node hexahedral elements, rather than the typical 8-node elements that can overestimate material stiffness, suffer from shear locking, and result in an inaccurate representation of stresses. While a variation of the SIMP method was employed that utilized hyperbolic tangent projection to enhance the discreteness of material distribution, the Kreisselmeire-Steinhauser function [104],

was also implemented to represent the buckling strength. The resulting optimized shapes expanded into the void areas external to the sphere, opened holes in the central adjoining regions, flattened out the spherical edges, and added external stiffening ridges. The geometry also somewhat resembled the Schwarz Primitive TPMS, minus the external ridges. However, those changes enlarged the thickness-to-area ratio to increase effective plate bending stiffness, and resulted in superior buckling strength. From their initial starting point, the optimized structure showed an improvement of 543% in buckling strength and 162% and 125% in bulk and Young's moduli, respectively.

Dong et. al. [105], proposed a design method that combined solid topology optimization with lattice support to create hybrid structures to be fabricated by AM. It was first determined through functional analysis which volumes of space would be solid versus lattice. Then, a uniform strut-based wireframe lattice was generated with an algorithm that modified it to remove short and disconnected ligaments. A model simulation with a Bidirectional Evolutionary Structural Optimization (BESO) method was employed to define strut thicknesses for the wireframe geometry, which were integrated with the solid portions to form the final model. The BESO method is a form of topology optimization that utilizes an FEA result to iteratively remove material from low-stress regions, and add material to high-stress regions, while keeping the total weight and volume constant. As displayed in Figure 2.23, the authors validated their method by 3-point bend experiments performed on samples produced via polymer FDM, and compared their hybrid architecture to those of a pure solid or pure lattice. They demonstrated that the hybrid design method produced the strongest structure with a 40% higher critical load when compared to the pure solid design.

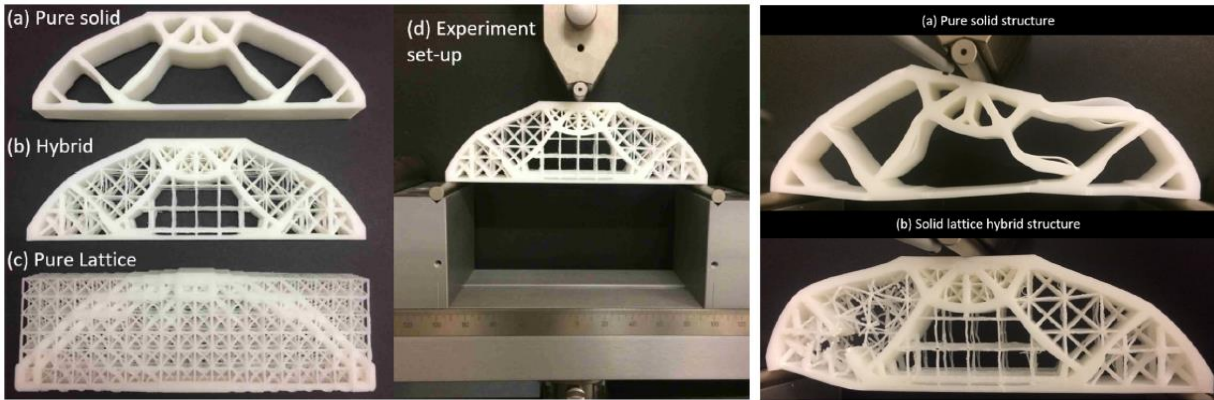


Figure 2.23. A pure solid SIMP topology optimization result compared with a hybrid and pure lattice result, demonstrating differences in failure modes [105].

Bici et. al. [106], presented a sandwich panel with a strut-type lattice core for a wing leading edge application. Aerodynamic loads, based on an aircraft during climb-out, were applied through an FEM analysis to investigate how structural element size and cell geometry would affect the stress-strain distribution. The authors created a design of experiments (DOE) to analyze the structural behavior based on variations in cell type (BCC or BCCZ), cell length (5–7 mm), strut radius (0.35–0.65 mm), and shell thickness (0.4–1.0 mm). An optimization was performed that used the Global Response Surface Method (GRSM) with two defined objectives: 1) mass minimization, and 2) maximize the first natural frequency, and two constraints that involved the strut and shell maximum stresses. Strut radius was found to have the most influence on stresses, while the BCCZ lattice outperformed the BCC at all cell lengths, though that also came with a weight penalty.

Günther et. al. [107], investigated shape optimization of TPMS lattice geometries after finding many of the conventional structural optimization techniques were limited to strut-based unit cell designs, and codes struggled with preserving biomimetic TPMS mesostructure for bone implants. The authors took advantage of the implicit solver contained within the nTopology platform, and

used the load cases important to implant applications such as compression, compression-shear, and shear, as design criteria. An example for the TPMS I-WP lattice is displayed in Figure 2.24. They employed the SIMP method on a 3^3 unit cell domain with a cell size of 3.33 mm, and a volume fraction of 0.25. The optimized TPMS skeletal Gyroid and I-WP lattice specimens were fabricated from Ti-42Nb using L-PBF. Following compression experiments, strength and stiffness was increased up to 61% and 80%, respectively, with the Gyroid geometry showing superior to the I-WP. This study indicated a strong potential for coupling shape optimization to TPMS morphologies.

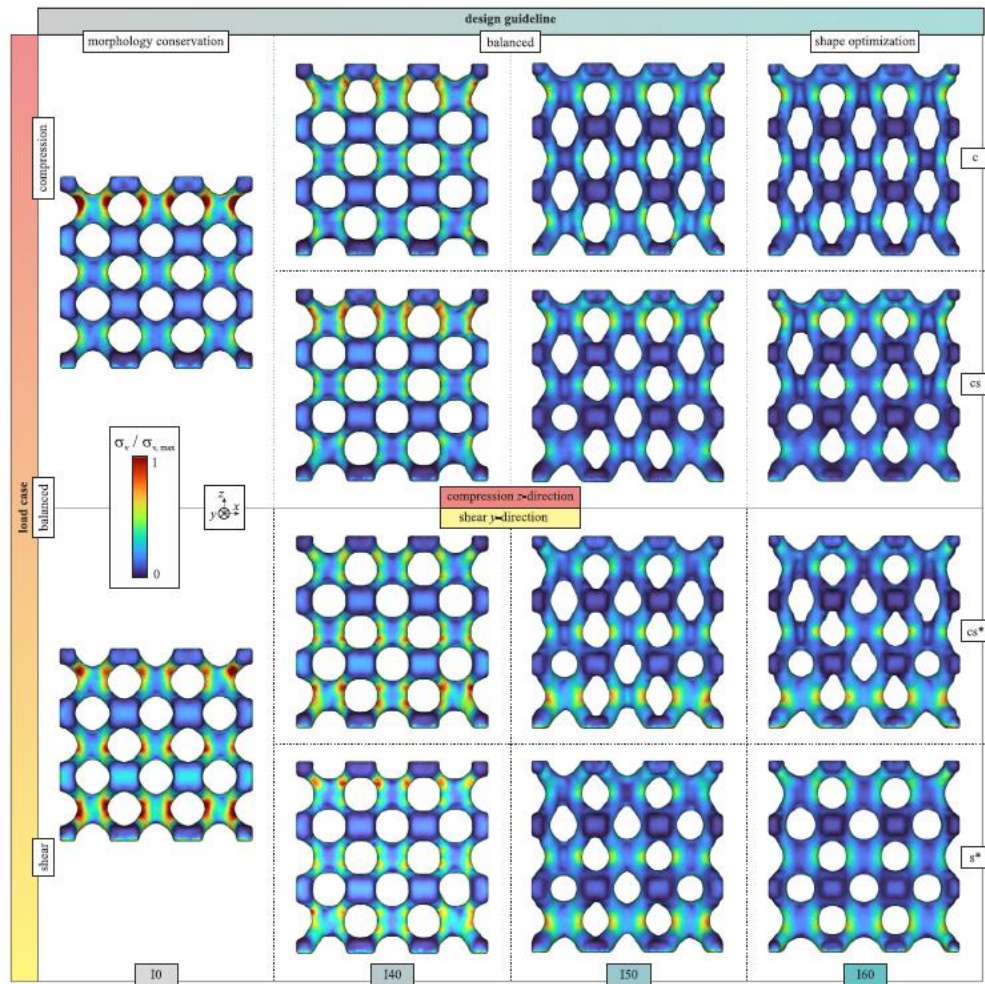


Figure 2.24. An example of von Mises stress plots for the I-WP lattice with different shape-optimized designs depending on the desired loading applications [107].

A machine learning method called a Back Propagation Neural Network (BPNN), was studied by Peng et. al. [108], to address challenges in the design of hybrid TPMS-based lattice structures. They discretized a two-dimensional 10 x 10 unit cell grid and assigned each cell randomly either a Gyroid or Primitive topology with a 25% relative density. Poisson's ratio and elastic modulus of the hybrid lattice structures was calculated based on the elastic tensor, according to homogenization theory. The network was trained on compression simulation finite element datasets, and it was able to predict the elastic modulus and Poisson's ratio of the hybrid TPMS lattices that were evaluated independently by FEA. The authors indicate that the BPNN predictions could be a potential tool for the design of lattice structures, though the numerical results were not validated with any empirical measurements.

Within available literature, the application of topology optimization to lattices has primarily been limited to the optimization of an individual unit cell, with periodic boundary conditions, and rarely at any macroscale end-use application [100]. The attempts at understanding functionally graded geometries of a non-uniform density, or the morphing of one topology to another, have been few. Typically, homogeneous methods were used for numerical analysis via FEM modeling to understand the behavior of a unit cell with periodic boundary conditions, and then those properties were applied to the macroscopic bulk material to avoid high computational costs. However, a number of experiments have shown that the post-yield deformation, and failure, to be non-periodic [69], so that behavior needs to be investigated using full models that contain enough unit cells to capture the effective properties and post-yield deformation. Often the differences between numerical predictions and experimental results were attributed to fabrication quality since defects significantly altered the overall mechanical behavior. There were no experiments conducted to examine tensile testing due to lack of ductility, nor has any fracture toughness results

of EB-PBF Ti-6Al-4V lattices been reported. Very little data to date is available in regards to the fatigue properties of lattice topologies, and all that are available were generated only in compression.

2.2.4 *Sandwich Structures*

In the aerospace industry, sandwich structures are common because of their low density (i.e. low weight), and high bending stiffness given their larger cross-sectional area. They consist of two face sheets with a core material bonded between them. For Example, sandwich panel construction as applied to composite structures utilizes face sheets made from carbon or glass fibers infused with a polymer matrix, while the core can be a different material, either metallic or non-metallic, often in a square or honeycomb cellular pattern [109]. Flexural stiffness of a sandwich panel is typically driven by the thickness of the core webs, the overall thickness of the panel, and the adhesion quality of the core to the face sheets. Sandwich panel structural integrity is lost when the core delaminates from the face sheets due to defects [110].

Within high temperature and high structurally loaded environments, fully metallic sandwich structures can be employed. Brazing, welding, or diffusion bonding are common processes used to join metal face sheets to a metallic core; although, the core-to-face sheet interface remains a vulnerable area. High stress concentrations at the interface, and the high probability for void defects, often lead to structural failure in bending. However, AM can provide the ability to fabricate a metal sandwich structure, in which the core and face sheets are monolithic, and fully integrated as a single component without a bondline. Being geometric complexity agnostic, PBF can produce core geometries that have been topology optimized for the intended application, and would be superior to traditional honeycomb.

One of the earliest monolithic metal sandwich structures that most resembles modern additively manufactured lattices was published in 2007 by Queheillalt et. al. [111], and shown in Figure 2.25. In that work, the authors extruded a 6061 aluminum alloy into a regular prismatic structure that had a core height of 19.1 mm, web thickness of 3.2 mm, and face sheet thickness of 5.2 mm. Along the length of the extruded triangular core webs, an alternating pattern of triangular-shaped electro-discharge machining (EDM) electrodes were inserted to create a pyramidal strut-type lattice sandwich panel with a relative density $\bar{\rho} = 6.5\%$. This article is notable because most sandwich structures typically suffer from node-failure as a result of a panel-to-core bond that is weaker than the parent material, yet quasi-static compression testing on this extruded-EDM lattice panel only demonstrated plastic yielding or inelastic buckling of the struts without any node failures. As such, the mechanical properties of the sandwich panel were governed only by the geometry, alloy, and method of loading. The non-dimensional compressive stiffness (Π) and shear stiffness (Γ) were predicted using Equations (2.50) and (2.51),

$$\Pi = \frac{E_c}{E_s \bar{\rho}} \quad (2.50)$$

$$\Gamma = \frac{G_c}{E_s \bar{\rho}} \quad (2.51)$$

where E_c and E_s are the Young's moduli of the core and solid parent alloy, respectively, and G_c is the core shear stiffness. This method demonstrated good agreement with the experimental stiffness measurements.

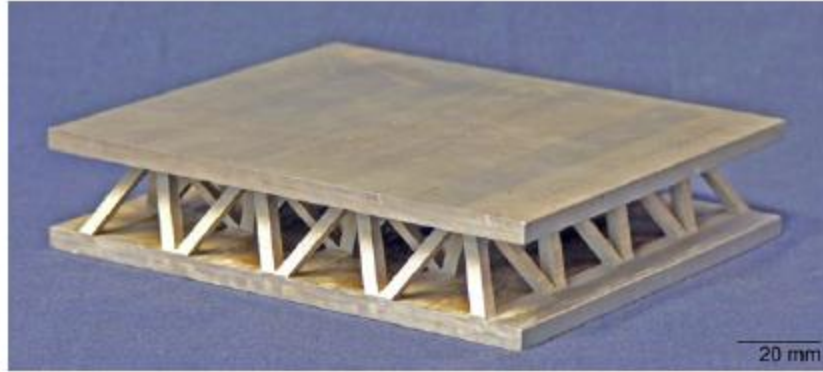


Figure 2.25. An aluminum extrusion following electro-discharge machining to create a pyramidal lattice Sandwich structure with a core relative density of 6.2% [111].

A two-dimensional honeycomb pattern, extruded into the 3rd dimension, is the most common form of core for sandwich panels in use by industries today; however, the effects of damage on crack propagation through the core remains an active area of research. By using additive manufacturing, authors Zhang, Deng, and Li [112], were able to conduct 3-point bend experiments on polymer printed honeycomb cores with designed-in defects. When cracks were placed at different orientation angles, they explored how propagation occurs based on the cell orientation, porosity, topological shape, relative density, loading speed, and various pre-fabricated crack parameters. From their results, there was a mix of plastic failures that were predominantly Mode-I open-type cracks, and some brittle failures consisting of Mode-I and a Mode-II shearing component. Two different configurations of core were shown to have the strongest crack suppression and load-bearing ability.

Oliviera et. al. [113] fabricated a cellular sandwich beam using the polymer FFF method of fabrication. The 13.5 mm unit cells were noted to mimic a sea urchin shape by cropping six sides of a hollow sphere, where the beam core was 1 unit cell thick, 3 unit cells wide, and 13 unit cells long. With factsheets that were 1.5 mm thick, they varied the core relative densities from 0.2 to

0.3 and evaluated the beams performance in 3-point bend experiments. In that work, the beam failures were driven by defects likely related to printed overhangs along a particular layer height. Therefore, simulations that did not account for defects, over predicted the stiffness K , the elastic slope of the load-displacement curve, by 30-40%.

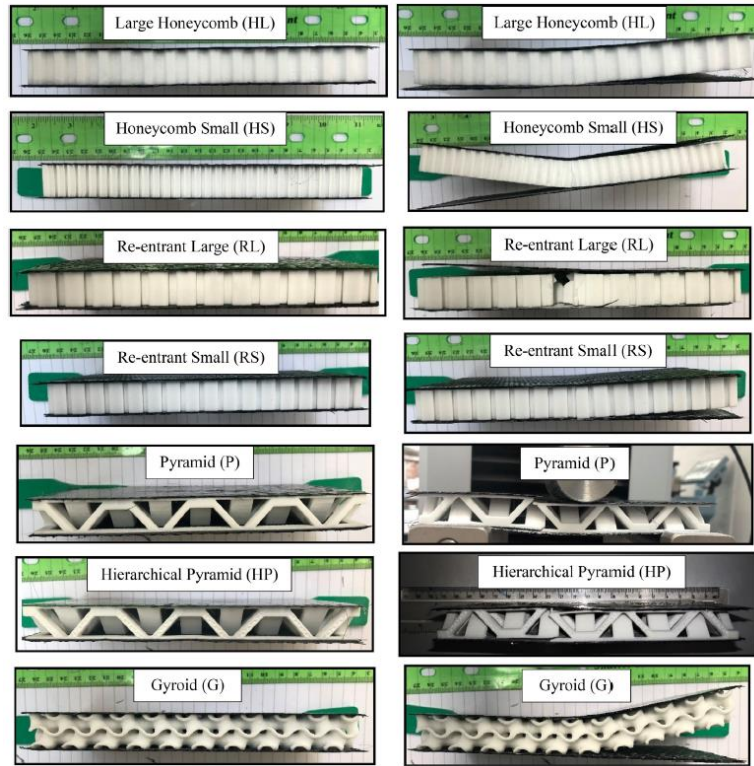


Figure 2.26. Composite sandwich structures with 3D printed core structures before (left) and after (right) 3-point bend experiments [114].

An investigation by Alshaer and Harland evaluated the strength and stiffness of sandwich beams constructed with 1 mm CFRP face sheets bonded to various PBF printed cores from Nylon PA12 [114]. The uniform core geometries, displayed in Figure 2.26, were a traditional hexagonal honeycomb with ‘small’ and ‘large’ cells, an auxetic re-entrant honeycomb with ‘small’ and ‘large’ cells, an octahedral lattice with a pyramid and hierarchical pyramid cells, and a TPMS Gyroid lattice. The thicknesses of the internal structures ranged from 0.5 mm to 1 mm, while the

overall beam core dimensions were 150 mm x 75 mm x 15 mm. Unit cell sizes varied depending on the particular geometry, and the Gyroid unit cell in particular was 10 mm³ with a sheet thickness of 1 mm.

Face sheets are considered “thin,” i.e., contributing less than 1% to the bending stiffness of the sandwich panel, if a ratio of core height h_c to face sheet thickness h_f satisfies Equation (2.52) [115].

$$\frac{h_c}{h_f} \geq 5.35 \quad (2.52)$$

Thicker skins reduce the core’s contribution to bending stiffness and the light-weight, porous functionality of a sandwich structure. Alshaer and Harland selected a thickness of 1 mm in order to reduce the likelihood of beam failure due to localized compression and skin wrinkling during 3-point bend experiments with a span of 110 mm. While such was the case, all the sandwich structures suffered from delamination between the core and face sheet materials along either the top or bottom surfaces. An important result was that the TPMS Gyroid structure was the only core that did not crack, and it demonstrated the highest normalized flexural strength, modulus, and stiffness (normalized values were obtained using the sample weight in Newtons).

In a study conducted by Monteiro et. al. [116], they investigated the 3-point flexural behavior of various strut-based lattice core topologies in a sandwich panel. Five lattice geometries, with a constant relative density of 30%, consisted of combinations of body and face centered parallelepiped crystalline structures. A single cell core height of 19.5 mm was selected, with a 1.25 mm thick face sheet, for the sample panels. They were constructed of polylactic acid PLA according to ASTM C393, which resulted in beam dimensions of 167 x 53 x 22 mm. Strut diameters varied depending on the lattice architecture, and the desired relative density, such that they ranged from 1.84 mm to 3.8 mm. An architecture that combined body and face centered

parallelepiped structures demonstrated the capability to compete with a traditional hexagonal core. One notable observation from this study was that failure of the specimens almost exclusively originated at the face sheet.

Scaling effects on the mechanical behavior of AM lattice structures was investigated by Aziz et. al. [117]. They employed the strut-based Body-Centered-Cubic (BCC) lattice in various sizes and demonstrated both compression, and 3-point flexural behaviors, via numerical, and experimental methods using a PLA polymer. From the flexure experiments, which had no corresponding numerical correlation, they showed that scale had no effect in the linear elastic region of the load-displacement curves, but in the non-linear regions the smallest scale sample behaviors appeared to be defect driven. For flexural strength comparisons, the authors simply used the maximum value of force in each case to show increasing values of 70 MPa to 82.5 MPa for the scale sizes of $\frac{1}{4}$ to 1. In compression, the lattice strength increased by 18% as edge length increased by a factor of 4. Overall, this study highlighted that there are scale effects to AM lattices to consider, and that sub-scale prototypes will likely represent a conservative estimate of the load-bearing characteristics of full-size components.

A comprehensive assessment of 10 different lattice strut-core types was conducted from a resin photopolymer manufactured via DLP (digital light projection) technology [118]. Part of the study involved sandwich beam specimens printed with face sheets and used in 3-point bend experiments, in addition to tensile and compression tests. One of the core types they employed was the Kelvin cell, which demonstrated a lower modulus than some of the other lattices. Additionally, it was noted that the Kelvin had a failure mode dissimilar to the others that exhibited sheet-core disjunction, much like a typical sandwich panel where the core and face sheet are bonded together.

Authors Bellini et. al. [119], evaluated the bending properties of a Ti-6Al-4V lattice sandwich beam manufactured via EB-PBF on an Arcam A2X. They used the approach of having the Materialise Magics software layout a uniform strut-type Octet core and then add the face sheets separately with an overlap sufficient to be printed as a single structure. This is significant because having the face sheets and core defined separately, the Arcam A2X can assign different process parameter themes to each, and in this case the authors selected the standard Melt theme for the face sheets, but a Net theme for the core. The Arcam coded Net theme uses the spot melting contour settings with a single high resolution sharp beam, and no multi-spot simultaneous melt pools. That is the manufacturer recommended theme for very fine features such as minimum thickness lattice structures.

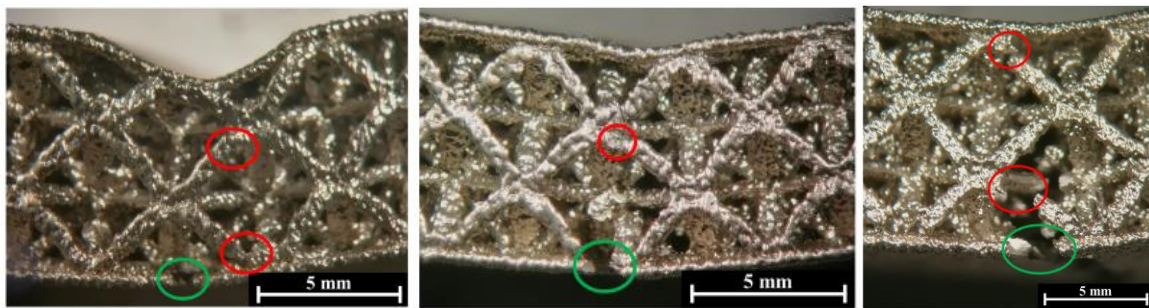


Figure 2.27. Images of failed 3-point bend EB-PBF titanium sandwich specimens with spans of 45 mm (left), 120 mm (middle), and 200 mm (right) [119].

After performing a manufacturability study, Bellini et. al. [119], determined the strut diameter to be 1 mm, and the Octet cell size to be 6 mm, with a 1.5 cell sandwich thickness and a 5-cell width. All of the sandwich beam specimens were manufactured vertically with different lengths of 70 mm, 165 mm, and 270 mm, and were subsequently flexural tested in 3-point bend. They looked at the failure modes of the various spans where shear stresses prevailed in the shorter specimens and normal stresses were predominate in the long ones. Local buckling was observed

adjacent the load input location with failure emanating from a crack in the lower face sheet mid-span that propagated vertically through the core. Their results were used in an attempt to calculate the load capacity for a generic bulk beam based on the measured flexural stiffness, though the outcome leaves this approach in question.

In another paper by Bellini et. al. [120], the uniform strut type 6 mm Octet unit cell, with a 1 mm strut diameter, was printed via EB-PBF into a thin titanium sandwich structure with a core thickness of 9 mm (1.5 unit cells) and 0.6 mm thick face sheets. Their sandwich specimens had a width of 30 mm, and varying spans of 45, 120, and 200 mm for 3-point bend experiments. By having three different span lengths (L), they could determine the bending stiffness (D) and shear rigidity (U) from the linear relation Equation (2.53),

$$\frac{w}{PL} = \frac{1}{48D}L^2 + \frac{1}{4U} \quad (2.53)$$

by plotting w/PL (mm/N) versus L^2 (mm²) in Figure 2.28, where P is the load at the end of the linear increase per unit width, and w is the displacement at failure, then obtaining a linear curve fit. Unfortunately, they failed to report the relative density of their structure, which is an important property to any lattice investigation.

Following the destructive 3-point bend experiments [120], a scanning electron microscope (SEM) was employed to examine the samples in regions of face sheet skin and strut fracture surfaces. While the images suffered from poor contrast and focus, and only displayed scale bars without any of the microscope settings, the various levels of magnification used ranged from 500 μm for images of a broken lattice strut, to 10 μm for the fracture surface. The images revealed the expected dimpling of a ductile fracture surface, as well as a lack of fusion pore at the fracture site with unmelted powder inside. They also showed broken struts that fractured in different ways and close ups of those fracture surfaces.

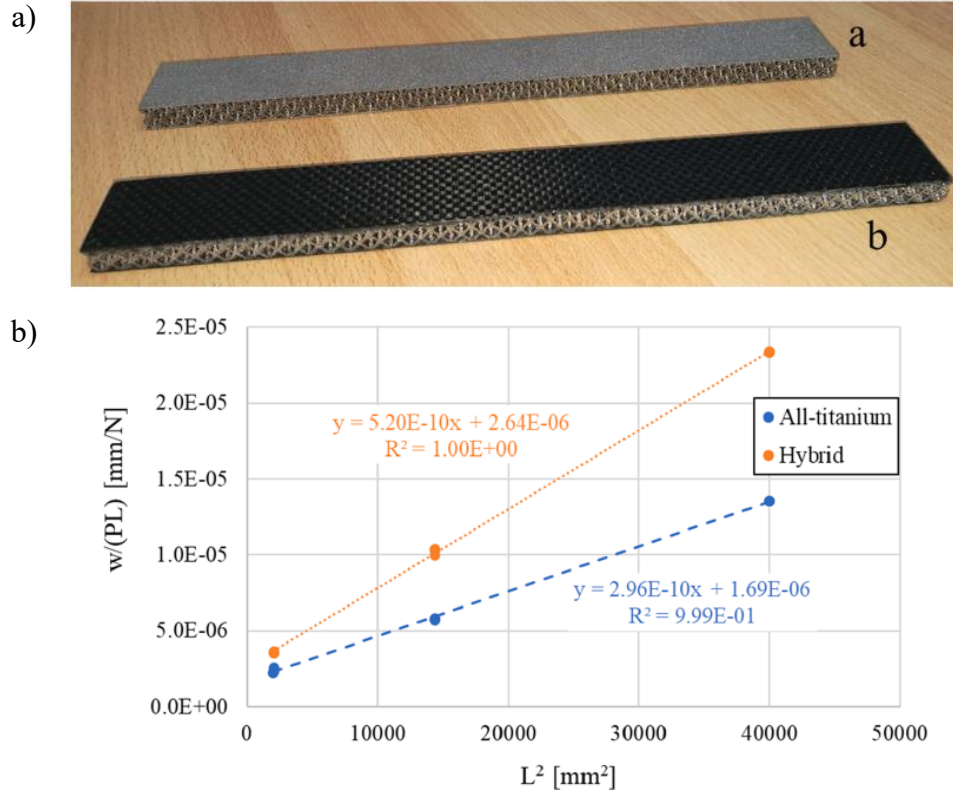


Figure 2.28. a) All titanium Octet lattice sandwich structure alongside a hybrid titanium lattice core with bonded CFRP face sheets, and b) a plot of experimental data with w/PL (mm/N) versus L^2 (mm²) [120].

For analysis, Bellini et. al. [120], chose to compare the bending stiffness of an all-titanium sandwich structure with a same thickness hybrid structure composed of a titanium lattice core bonded to CFRP face sheets. No other predictive analytical or numerical methods were employed. Where the all-titanium structures showed a much higher bending stiffness, on the order of 40% more, the authors noted that the weight was as much as 20% more. They proceeded to then use the following relation Equation (2.54) from ASTM D7250

$$D = \frac{Eb(h^3 - h_c^3)}{12} \quad (2.54)$$

to calculate what the bending stiffness would be if the weight delta was neutralized by adding additional CFRP face sheet thickness. That resulted in a bending stiffness more than 2.5 times higher than the all-titanium structure. It is not clear why this was a good comparison, because adding thickness will naturally increase the cross-sectional area, and the bending stiffness, which is the whole point of a sandwich structure. Additionally, the amount of skin thickness was more than doubled to 1.5 mm. If the authors wanted to show a weight neutral comparison, they would have traded core thickness for laminate thickness, as well as breakdown the stiffness contributions between laminate and core. At 3 mm of total face sheet thickness compared to a core of 9 mm, the face sheets now take up a significant percentage of the total panel thickness.

Zhou et al. [81], used the elastic deformation constitutive relation to analytically determine whether their four hierarchical level unit cells were bending-dominated or stretch-dominated. They tested their theory numerically with good agreement, -7.2% to 8% error, but saw less agreement experimentally. Lattice sandwich specimens were fabricated from AlSi10Mg via L-PBF, with struts 0.6 mm in diameter, face sheets 0.6 mm thick, and the specimens contained five layers of unit cells between the face sheets. The results of the compression experiments are shown in Figure 2.29 where the first, second, and third hierarchical levels demonstrated a bending-dominated behavior, and the fourth clearly resembled a stretch dominated behavior. While the authors did not address this discrepancy directly, they did report that the strut diameters varied by $\pm 10\%$, and the elastic modulus of the AM lattices had an error range of -18% to 23% from the design value which they described as reasonable.

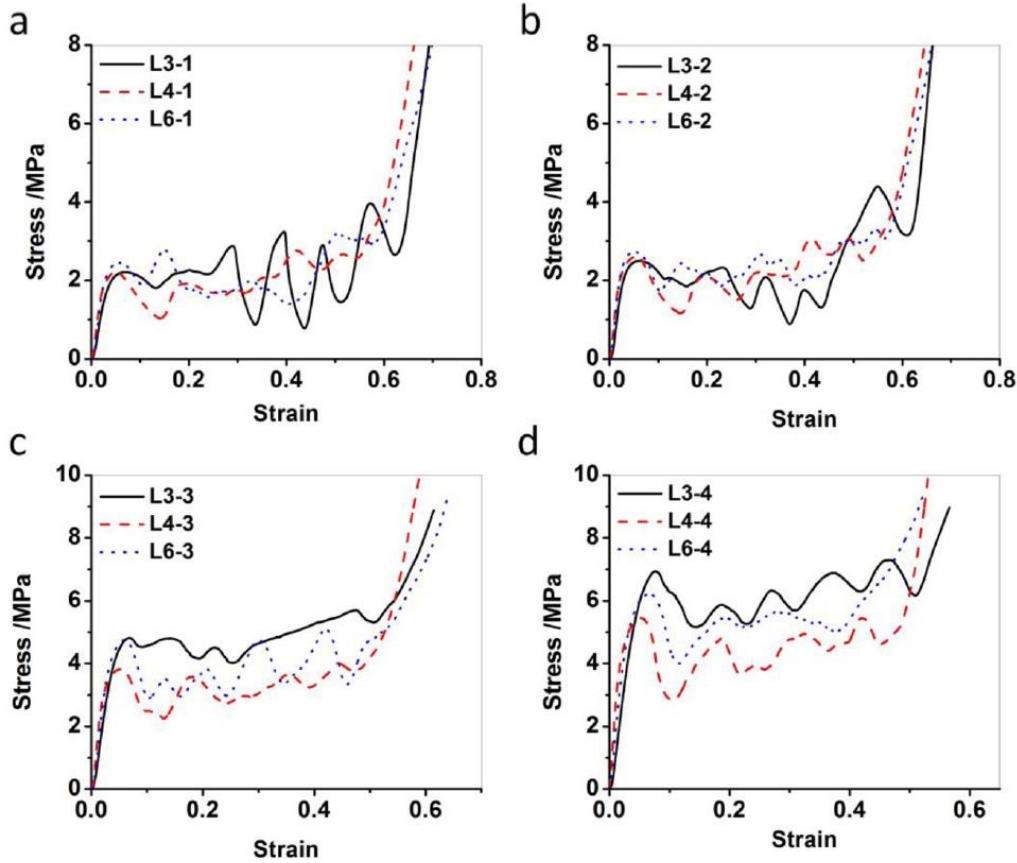


Figure 2.29. (a) First hierarchical level, (b) second hierarchical level, (c) third hierarchical level, and (d) fourth hierarchical level self-supporting lattice structures experimental stress–strain curves [81].

A novel L-PBF sandwich panel was proposed by Ferro et. al. to function as a structural heat exchanger for an aircraft application [121]. The hollow AlSi10Mg panel would serve as an aerodynamic surface, while a trabecular lattice core would allow hot air to circulate through the structure for de-icing purposes. The trabecular core design was selected for its high specific surface area for optimum heat exchanging capability, but this study focused on the compressive strength of the Body-Centered Cubic (BCC) geometry and one with added vertical struts (BCCZ). With a strut size of about 1 mm and a unit cell size of 4 to 5 mm, the lattices were evaluated based on their specific elastic modulus (E/ρ), specific yield stress ($\sigma_{0.2}/\rho$), and specific maximum stress

(σ_{max}/ρ). Their results showcased buckling failure modes of the struts, with those in the vertical direction obviously performing better in compression. Also, those with the thicker struts and smaller cells, i.e. higher density, demonstrated superior strength, though there was no reflection on how well these configurations actually performed as heat exchangers. Increased density would hinder fluid flow through the structure and provide a pressure drop penalty such that a balance between strength and heat transfer would be required.

Three-point bend experiments were used by Daynes et. al. to test sandwich structures comprised of BCC lattice cells [122]. They employed two functionally-graded core designs that were optimized based on tailoring the lattice strut diameter and cell special size. These methods increased the stiffness and strength when compared to a uniform topology core, although the specially graded lattice performed best because it had a greater number of design variables to optimize. Added degrees of design freedom were then introduced to a novel isostatic line method that functionally graded the lattice size, aspect ratio, and orientation, to align load-bearing elements with the principal stresses. Isostatic lines describe the maximum and minimum principal stress trajectories based on a FEA of a density-optimized lattice core. However, the actual strength of the structures fell short of the predicted strengths due to localized buckling and core crush that was beyond the capability of a linear static analysis.

Sandwich lattice beams have also been looked at as a means to control wave propagation. Guo et al. [123], investigated the band structure of flexural waves with an hourglass lattice truss core as a metamaterial sandwich beam. The lattice core was modeled with lumped masses in a series of local resonators in an effort to provide a roadmap for modeling lattice sandwich beams with complex core structures to control wave propagation. In another effort by Jiansheng Pan et al. [124], they looked into the flexural properties of additively manufactured TPMS lattices as

hinges. Their study used FEA to analyze the compliance of the Primitive, Gyroid, Diamond, and I-WP, where they found non-uniform deformation of the Diamond and Gyroid, and the I-WP to have less compliance than the Primitive. They then used the Primitive TPMS lattice for flexural experiments with its advantages in large compliance and compliance ratio, and looked at ways to improve upon it.

While sandwich structures may not be a new phenomenon, it is clear from this research that information on additively manufactured metallic sandwich structures, that can take advantage of the ability to spatially optimize the core geometry, is lacking. Much of the investigations on latticed sandwich cores was actually conducted using polymer additive processes due to their low cost and wide availability. Additionally, many researchers chose to conduct flexural experiments in 3-point bend, which is more applicable to homogeneous and isotropic materials, such as a piece of metal bar stock. However, 4-point bend experiments provide a better representation of the flexure behavior for a sandwich structure, because it creates a more realistic combined loading scenario simulating real-world conditions. Between the two load points, the material faces a more uniform stress distribution caused by a constant moment, which can expose flaws that may not have been apparent in a point load scenario. Therefore, a significant knowledge gap exists around the flexural behavior of metal additive sandwich structures with optimized lattice cores.

2.3 MATERIALS

In general, metal alloys that are easily weldable are best suited for EB-PBF, due to the similarities in industrial fusion welding and the fusion of powder particles. Powder metallurgy is an industry that has been around much longer than PBF additive manufacturing so the processes for making the powder feedstock are well established. Typically, metal powders are pre-alloyed and fabricated from melting and atomizing other forms of feedstock like ingot or bar. Since the

material used exclusively for this research was the titanium alloy Ti-6Al-4V, this section will provide the relevant background on this material, its α/β phase microstructure, and how it has been used in PBF research.

2.3.1 *Constitutive Modeling*

Mechanical properties, such as yield strength, ultimate strength, and elastic modulus, of a homogeneous material are typically determined through tensile experiments where the material is pulled apart while measuring the force and displacement. The measurements are then used to calculate the stress and strain using Hooke's law, and then from the stress-strain curves, material properties such as elastic modulus, yield strength, and ultimate strength can be derived. The elastic modulus being the initial linear slope of the elastic region of the stress-strain curve. Once the material hits the yield strength point, damage sets in and it begins to permanently plastically deform.

For a brittle material, the deformation beyond yield will be very short, showing little to no elongation prior to fracture. The non-linear plastic region of Ti-6Al-4V typically shows large strain deformation, indicating that it is a ductile material. Some materials will continue to show some amount of increase in stress as strain increases, which is referred to as post-yield, or strain hardening, while other materials will show a post yield softening as stress decreases. In the case of Ti-6Al-4V, this material demonstrates a plateau in stress as strain continues to increase to failure, and this behavior is considered to be perfectly plastic. While Hooke's law provides a way to analytically describe the linear-static elastic behavior, it does not extend beyond yield into the plastic regime. Therefore, researchers have developed additional models such as GTN or CDM to provide analytical predictive capabilities for materials.

In classical plasticity, there is an assumption that yielding is independent of hydrostatic stress [125]. The Gurson-Tvergaard (GT) model analyzes plastic flow in a porous medium by introducing voids that influence the global flow behavior. This adds a strain-softening effect to conventional plasticity theory, and it contains a failure criterion where plastic instability that produces a band of local deformation causes a ductile fracture. The GT model, shown in Equation (2.55), originates from a rigid-plastic limit analysis. The solid has a smeared volume fraction f of spherical voids, and the yield surface and flow potential g , is also a function of the macroscopic von Mises stress σ_e , the macroscopic mean stress σ_m , and the flow stress for the matrix material of the cell $\bar{\sigma}$.

$$g = \left(\frac{\sigma_e}{\bar{\sigma}}\right)^2 + 2q_1f \cosh\left(\frac{3q_2\sigma_m}{2\bar{\sigma}}\right) - (1 + q_3f^2) = 0 \quad (2.55)$$

This model contains the constants $q_1 = 1.5$, $q_2 = 1.0$, and $q_3 = q_1^2$, which are typical values used for metals. The von Mises yield surface for an incompressible material can be obtained by setting $f = 0$.

This model was further modified by Tvergaard and Needleman (GTN) to include void coalescence by replacing f with f^* , an effective volume fraction that amplifies the hydrostatic stress effect when $f > f_c$, the critical void volume fraction where stress carrying capacity begins to decay more rapidly. The relation shown in Equation (2.56) attempts to mimic how voids grow rapidly in real materials, shown in Figure 2.30, when the void fractions exceed 10-20%.

$$f^* = \begin{cases} f & \text{for } f \leq f_c \\ f_c - \frac{f_u^* - f_c}{f_F - f_c}(f - f_c) & \text{for } f > f_c \end{cases} \quad (2.56)$$

Equation (2.56) also contains the fitting parameters of the final void volume fraction f_F where there is a complete loss of stress carrying capacity (material fracture), and $f_u^* = 1/q_1$ (value of f^*

at zero stress state). Tvergaard and Needleman assumed $f_F = 0.25$ and $f_c = 0.15$ [126]; however, some of these parameters are not particularly useful if failure is assumed when $f > f_c$.

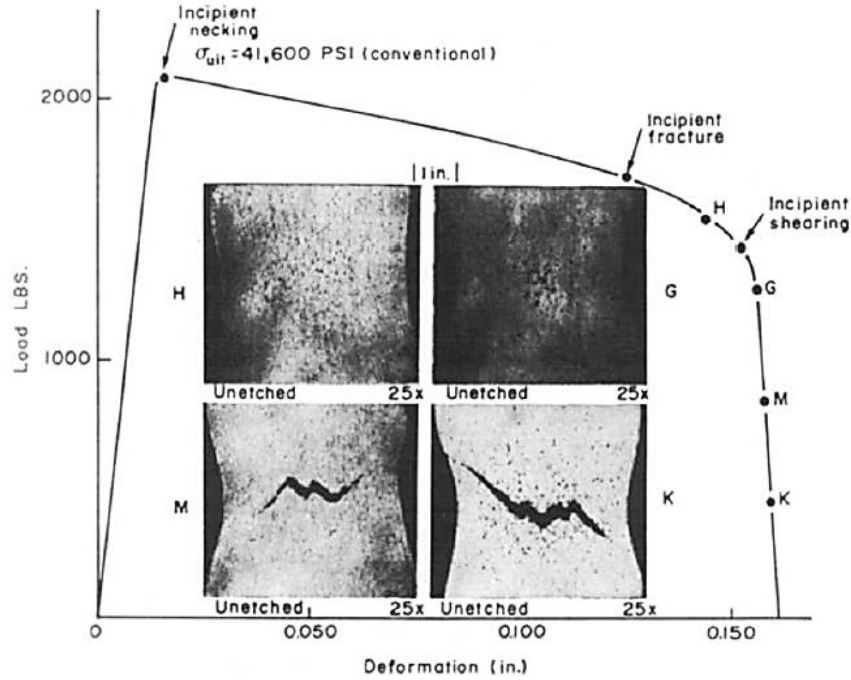


Figure 2.30. Failure of a round tensile specimen is it relates to the load-deformation curve (originally from Bluhm and Morrissey) [126].

A constitutive modeling and numerical simulation study was performed by Yang et. al. [127] on the ductile fracture behavior of L-PBF Ti-6Al-4V using a GTN model modified by Jiang [128]. Their J-modified GTN model introduced a continuum shear damage variable to make the model dual variable in order to address void shear failure at low and negative stress, in addition to the already accounted for fracture behavior at medium to high stress. In this way, the applicability of the GTN model was extended to a wider range of stress states, the J-modified GTN model is shown in Equations (2.57), (2.58), and (2.59).

$$\Phi = \frac{q^2}{\bar{\sigma}_m^2(1-D^*)^2} + \left\{ 2q_1 f^* \cosh\left(-\frac{3q_2}{2} \frac{p}{\bar{\sigma}_m(1-D^*)}\right) - 1 - (q_1 f^*)^2 \right\} = 0 \quad (2.57)$$

$$f^* = \begin{cases} f & f \leq f_c \\ f_c + K_f(f - f_c) & f_c < f < f_F \\ f_u^* & f \geq f_F \end{cases} \quad (2.58)$$

$$D^* = \begin{cases} D & D \leq D_c \\ D_c + K_D(D - D_c) & D_c < D < D_F \\ D_u^* & D \geq D_F \end{cases} \quad (2.59)$$

The variables $K_f = (f_u^* - f_c)/(f_F - f_c)$ and $K_D = (1 - D_c)/(D_F - D_c)$, where D is the current shear damage of the material, D_c is the critical shear damage, and $D_u^* = 1$ which is the value of D^* at zero stress. As follows, the damage variable f^* models the rapid loss of stress carrying capability as voids coalesce, and D^* denotes the stage of coalescence in micro-shear bands when the critical condition is reached.

A challenge of using the GTN model in practicality has always been that it is difficult to determine the numerous parameters and derivatives. A unified method of calibrating the explicit parameters for the model has yet to be established. However, Yang et. al. [127], proposed a calibration strategy that was based on a finite element inverse analysis and used results of tensile and compression experiments. Their parameters were then validated by simulating additional experiments where they observed that the void fraction variable controlled the void growth mechanism of the fibrous zone failure, and the shear damage variable enabled the void shear mechanism of shear fracture.

The study conducted by Sérgio et. al. [129], coupled the GTN damage model with the typical da/dN - ΔK curves to build a numerical model to predict fatigue crack growth (FCG). Crack propagation occurs by node release as cyclic plastic strain accumulates to critical values, and the GTN model was employed to describe the material degradation as micro-voids grew and coalesced at the crack tip. It was determined that the numerical predictions, without the inclusion of micro-void growth from the GTN model, had underestimated the slope of the da/dN - ΔK curves from

experimental data shown in Figure 2.31. The authors reported that the Paris-Erdogan law parameter m was 3.62, which was much higher than the value obtained without GTN of $m = 2.62$. The addition of GTN provided an anti-clockwise rotation of the curve to more closely match the experimental results and produced a slope value of $m = 3.36$.

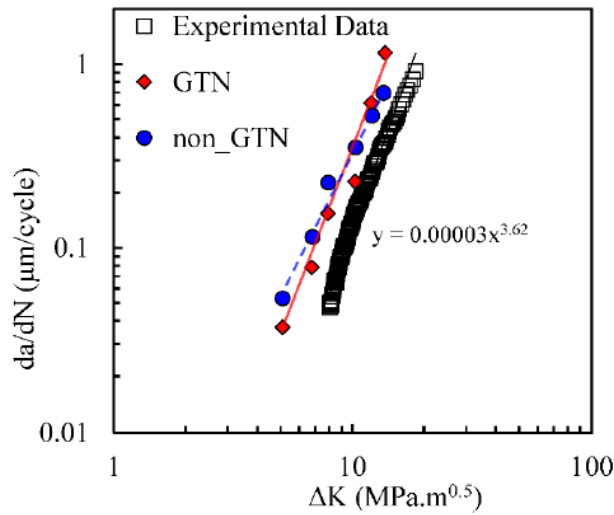


Figure 2.31. Log-log plot of da/dN - ΔK curves for the analytical models, with and without GTN, compared with experimental data [129].

Another popular analytical damage model called Continuum Damage Mechanics (CDM) theory was established by Kachanov [130], which defined effective stress. CDM models simulate initial damage, which grows after reaching a plastic deformation threshold strain, and then proceeds to failure. Within the framework of CDM, Lemaitre [131], considered how stress tri-axiality affected damage growth, plastic deformation, and fracture. In a study conducted by Rokhgireh et. al. [132], the authors used CDM (Lemaitre) and GTN failure criteria to model the tensile behavior of L-PBF manufactured specimens of different materials, including Ti-6Al-4V. CDM starts with a plasticity model, with von-Mises yield criterion (F), and an isotropic strain hardening function (R), shown in Equations (2.60) and (2.61) respectively.

$$F = \frac{\sigma^{eq}}{1-D} - (\sigma_y + R) = 0 \quad (2.60)$$

$$R = R_\infty [1 - e^{-\gamma r}] \quad (2.61)$$

$$D = \frac{A_D}{A} \quad (2.62)$$

Here, σ_y is the yield stress, with σ^{eq} representing the equivalent von-Mises stress, R_∞ is the stabilized value of isotropic hardening, while γ controls the stabilization rate with respect to accumulated plastic strain r . The isotropic damage parameter D in Equation (2.62) is the ratio between the surface area of voids and micro-cracks A_D , and A the total surface area. Equivalent von-Mises stress is a function of the deviatoric stress tensor components S_{ij} as shown in Equation (2.63).

$$\sigma^{eq} = \left(\frac{3}{2} S_{ij} S_{ij} \right)^{1/2} \quad (2.63)$$

In order to represent the inherent defects in L-PBF produced material, Rokhgireh et. al. modified the isotropic CDM (Lamaitre) by adding an initial non-zero damage parameter D_0 . Conditionally, if accumulated plastic strain is less than the strain at which damage occurs, $r < \varepsilon_D$, damage does not change and the change in stress is governed by Equations (2.64), (2.65), and (2.66),

$$d\sigma_{ij} = (1 - D_0) C_{ijkl}^{EL} (d\varepsilon_{kl} - d\varepsilon_{kl}^{PL}) \quad (2.64)$$

$$C_{ijkl} = \lambda \delta_{ij} \delta_{kl} + 2\mu \delta_{ik} \delta_{jl} \quad (2.65)$$

$$d\varepsilon_{ij}^{PL} = \frac{3}{2(1-D)} \frac{3\mu S_{kl} S_{ij} d\varepsilon_{ij}}{(\sigma^{eq})^2 [3\mu/(1-D) + R_\infty \gamma e^{-\gamma r}]} \quad (2.66)$$

where λ and μ are Lamé's constants, δ_{ij} is Kronecker's delta, and $d\varepsilon_{ij}^{PL}$ is the plastic strain tensor. However, if $r > \varepsilon_D$, then the damage parameter increases according to Lemaitre's model in Equations (2.67) and (2.68),

$$dD = \left(\frac{-Y}{S_0} \right)^{S_1} \frac{dr}{1-D} \quad (2.67)$$

$$dr = \left(\frac{2}{3} d\varepsilon_{ij}^{PL} d\varepsilon_{ij}^{PL} \right)^{1/2} \quad (2.68)$$

where S_0 and S_1 are model material parameters and Y is the energy release rate.

The yield criteria for the GTN model was obtained by substituting σ_y for $\bar{\sigma}$ in Equation (2.55), and defining $\sigma_m = \sigma_{kk}/3$ as the mean (hydrostatic) stress. An initial void volume fraction was described by f_0 and it was stated that Tvergaard and Needleman assumed $f_F = 0.25$ and $f_c = 0.15$. Then, in order to combine the CDM and GTN models, both were set as functions of void volume fraction f by making the CDM damage parameter D a function of f through a geometric relation described by Equations (2.69) and (2.70).

$$f = \frac{4/3 \pi R^3}{L^3} \quad (2.69)$$

$$D = \frac{\pi R^2}{L^2} = \left[\frac{3\pi^{1/2}}{4} f \right]^{2/3} \quad (2.70)$$

In this way, properties of the material, such as Young's Modulus E , yield stress σ_y , R_∞ , and ε_D , would become functions of the initial damage parameter D_0 as described by Equations (2.71), (2.72), (2.73), and (2.74).

$$E = (1 - D_0)E_0 \quad (2.71)$$

$$\sigma_y = (1 - D_0) \quad (2.72)$$

$$R_{\infty} = \frac{R_{\infty}^0}{(1 - D_0)} \quad (2.73)$$

$$\varepsilon_D = -\frac{1}{\gamma} \ln(D_0) \quad (2.74)$$

The results presented by Rokhgireh et. al. [132], demonstrated that unlike CDM in Figure 2.32a, the GTN model was not able to adequately predict failure since the elastic modulus and failure strain did not depend on initial porosities. They also found that there were size-dependent effects, shown in Figure 2.32b, such that as the sample diameter increases, the initial defects D_0 decreases. This makes sense, because in small diameters, the voids contained in PBF are going to represent a larger portion of the cross-sectional area than in larger scale bulk material. Essentially, as sample diameter increases, the number of defects will increase, but their size will remain small and not grow in relation to the sample size. Therefore, while the material volume is increasing at an exponential rate, the void volume will only increase in a more linear fashion causing D_0 to decrease.

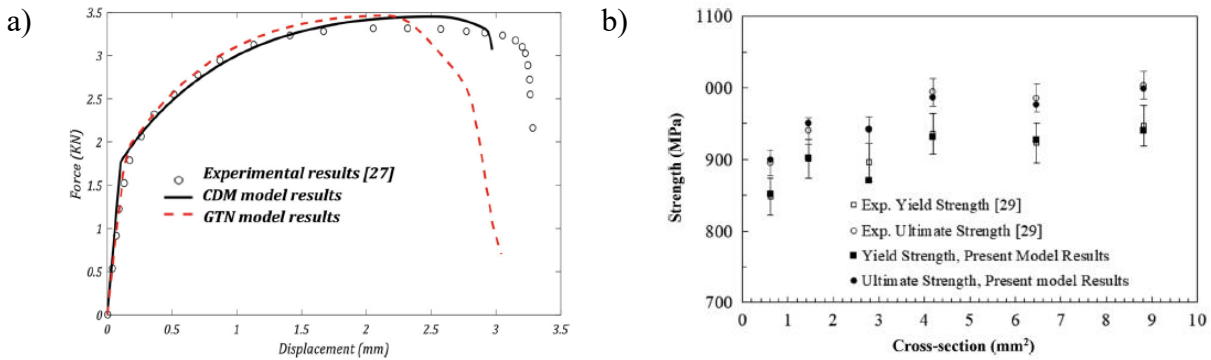


Figure 2.32. a) Comparison of predicted behavior between the GTN and CDM constitutive analytical models, and b) CDM predictions compared with experimental results of Ti-6Al-4V as they relate to sample cross-sectional area [132].

Homogenization models are where material properties are determined at a local micro or meso scale, and then applied globally to a macro scale bulk material. This approach is common for

composite materials and sandwich panels with honeycomb core. However, with the rise in interest of 3D printed architected lattices, Molavitabrizi et. al. [133], found a need for a 3D homogenization model addressing voided materials with complex geometries. They took a strain-gradient homogenization theory approach to develop a periodic boundary condition model that does not assign a material to void regions in order to simplify the analysis. The micro-scale and macro-scale effects were bridged using Hill-Mandel lemma, with the finite element method, to study the model's behavior on 2D square, and 3D cubic lattice materials. Model validation was derived from 3-point bend experiments using SLS PA-12 nylon beams and compared the predicted versus measured reaction forces. A difference of only 7% was found between the full-field simulations and the experiments, which was attributed to issues such as print direction related mechanical properties, and shape distortion from the fabrication process. This study points out that more investigation is needed into a suitable continuum theory, based on the topology of an architected material and the loading conditions.

Modeling the fatigue behavior of Ti-6Al-4V fabricated by PBF has been the subject of recent research as well. For example, Le et. al. [134], presented their work with ISO 2072:2010 fatigue coupons produced using the L-PBF process in order to characterize the behavior relative the process induced porosity. The microstructure of the produced specimens under optical microscopy matched the expected columnar grain anisotropy associated with the build direction. Kitagawa-Takahashi diagrams showed that pore size controls the fatigue resistance, and that the oriented microstructure did not have a pronounced effect. The authors employed two commonly known fracture mechanics approaches, the Paris Law, and the Canton model, to predict fatigue life.

The Paris Law [135], [136], for modeling crack growth and fatigue life is shown in Equations (2.75) and (2.76), where ΔK is the stress intensity factor (SIF) and Y equals 0.65 for surface pores.

$$\frac{da}{dN} = C(\Delta K)^n \quad (2.75)$$

$$\Delta K = Y\Delta\sigma\sqrt{\pi\sqrt{area}} \quad (2.76)$$

The Canton model [137], [138], [139], assumes a small crack growth rate is a function of macroscopic maximum strain ε_{max} , stress amplitude σ_a , and material yield strength σ_y , demonstrated by Equation (2.77), with a as the defect size.

$$\frac{da}{dN} = D \left(\varepsilon_{max} \left(\frac{\sigma_a}{\sigma_y} \right)^s a \right)^t \quad (2.77)$$

The two models were compared in Equations (2.78), (2.79), (2.80), and (2.81), using Murakami's parameter $\sqrt{area}$ to define the defect size.

$$\frac{d(\sqrt{area})}{dN} = C' \left(\sigma_{max} (\sqrt{area})^{1/2} \right)^n \quad (2.78)$$

$$C' = C(Y(1-R))^n \pi^{n/2} \quad (2.79)$$

$$\frac{d(\sqrt{area})}{dN} = D' \left(\sigma_{max} (\sqrt{area})^{1/2s} \right)^{t'} \quad (2.80)$$

$$D' = D \left(\frac{(1-R)}{2E\sigma_y} \right)^{st}, \quad t' = 2st \quad (2.81)$$

In this way, it was shown that the Canton model is an extension of the Paris law, as when $s = 1$ the Canton model simplifies to the Paris law. As shown in Figure 2.33, Le et. al., were able to get better agreement with the Canton approach, after correlation with experimental results and assisted by the Kitagawa-Takahashi diagrams.

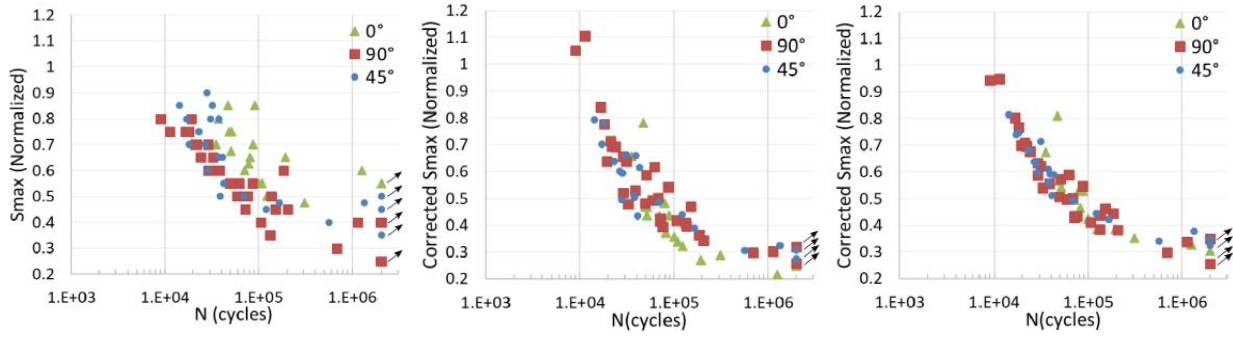


Figure 2.33. S-N curve comparison between the raw data (left), Paris Law corrected-stress data (middle), and the Canton corrected-stress data [134].

A study published by Luan, Chen, and Gaitanaros [140], followed a global energy-based approach to investigate the fracture properties of strut-based lattice materials. Classical linear elastic fracture mechanics assumed an asymptotic stress field ahead of the crack, and that fracture occurred when the stress at the strut ahead of the crack tip reached the parent solid material fracture strength. The energy-based method calculates the energy release rate as the crack propagates (a measure of toughness), rather than using a critical value for the stress intensity factor. However, determining the fracture energy and crack propagation path a-priori are the main challenges to that approach, which the authors addressed through theory, simulations, and experiments.

As displayed in Figure 2.34, the authors employed 2D triangular and honeycomb lattice structures in uniaxial tension, with various crack sizes initiated in the 90° or 30° directions relative to the vertical loading direction [140]. The energy release rate of the triangular G_T and lattice was calculated via Equation (2.82)

$$G_T = \eta_T \frac{\sigma_0^2 a}{E_T^*} \quad (2.82)$$

where σ_0 is the stress applied at the boundary, a is the crack length, η_T is determined by coefficients that when converged, $\eta_T = \pi$. Thus, a sufficiently large plate of triangular lattice with

a very small crack would behave similarly to a monolithic material; however, this type of lattice is one of the most stretching-dominated topologies and generally not representative for all 2D lattices.

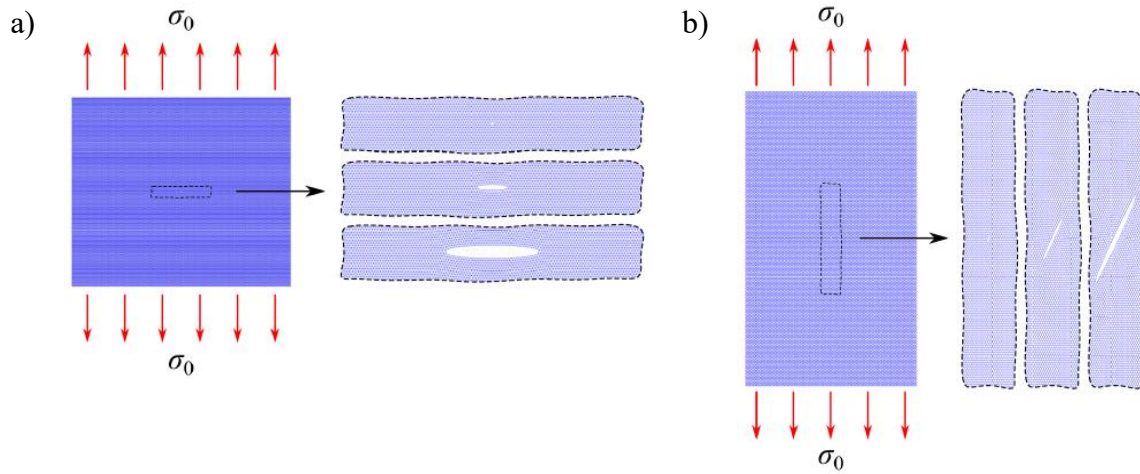


Figure 2.34. Crack formation under uniaxial tension in a 2D a) triangular lattice, and b) honeycomb [140].

That was shown to be the case for the honeycomb lattice with the 30° incline crack where the energy release rate G_H was the same as Equation (2.82), but η_H converged to a different value that made its behavior bending-dominated [140]. Analytically, it was shown that the honeycomb lattice energy release rate was much more sensitive to its relative density than the triangular lattice due a cubic scaling of the effective elastic modulus with respect to density. After comparisons with FEM simulations and experimental findings, the energy-based approach provided similar results to a stress-based local approach. The framework presented by the authors, does not rely on assumed restrictions from classical linear elasticity, and highlighted the complexity of lattice-based fracture behavior.

Much remains to be understood when it comes to the mechanical behavior of materials and lattice structures. While existing models may work well in some cases, they are often plagued

with numerous assumptions or parameters that have to be empirically derived such that their applications remain limited. The popular method of homogenization, and many other models, quickly becomes unusable when printed materials can often have strong build direction anisotropy, or lattices structures are topology optimized and tailored to the application. Therefore, more robust analytical models are needed to better address a future that contains architected materials.

2.3.2 *Metal Powder Bed Fusion*

The benefit of an established powder feedstock industry is that producers operate to standards and have controlled processes and quality assurance. Different methods for powder production result in different powder characteristics such as size, morphology, flowability, and chemical composition. For PBF in general, the feedstock commonly used is highly spherical, and produced through various atomization methods [39]. The sphericity helps to facilitate particle flow within the machine for uniform layer distribution, as well as provide a homogeneous packing density. Chemistry requirements for Ti-6Al-4V powder are laid out in ASTM F3001 and the typical particle size for EB-PBF ranges from 45-105 μm with an average around 75 μm . For contrast, L-PBF utilizes a finer particle size distribution that averages around 36 μm , as shown in Figure 2.35 [141].

In Laser and EB-PBF, when the beam energy is removed and the temperature of the melt pool drops quickly, it causes the surface tension of the melt pool to increase. So long as the surface tension is higher than the metal vapor pressure, then vapor bubbles are trapped in the melt pool to form porosity defects. Oscillating mirrors, driven by a step-wise motor, causes the laser beam spot movement in L-PBF to be discontinuous, but the EB-PBF process generates a continuous beam, which results in a wider melt pool than the one produced by a laser. The smaller melt pool has a greater surface tension, and thus more defects tend to form. Pore defects and balling can also be

created with excessive energy density, such as a high beam power and low scan speed, which induces melt pool oscillations and Marangoni effect [142].

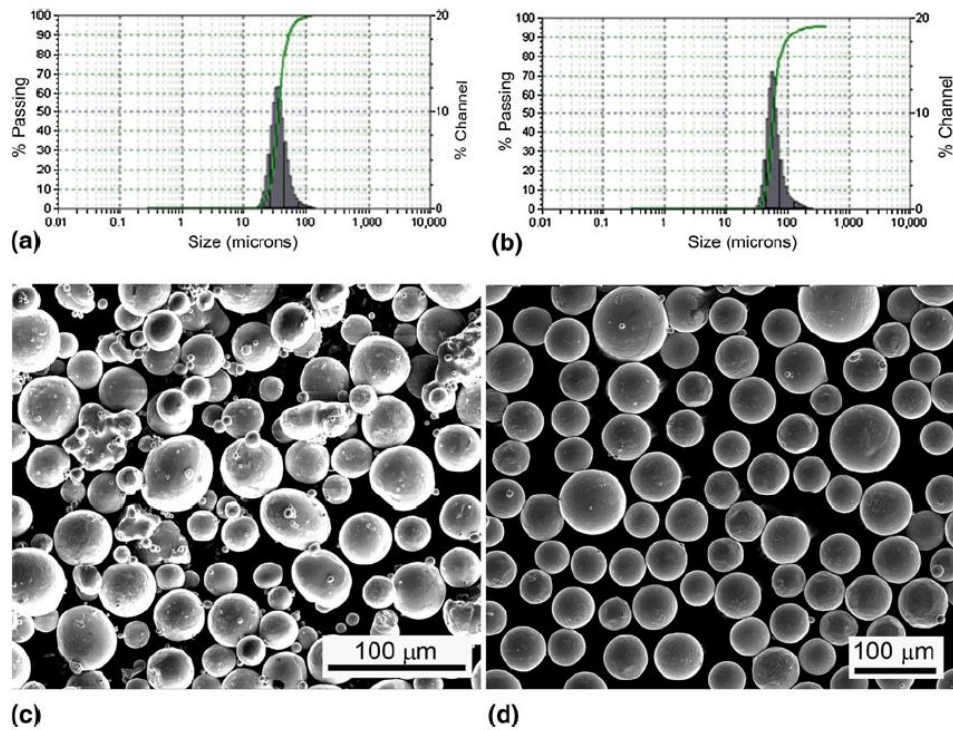


Figure 2.35. Particle size distributions and SEM images of (a) (c) L-PBF powder, and (b) (d) EB-PBF powder [141].

A review conducted by Sola and Nouri [143], revealed recurring mechanisms for microstructural pores that are found in PBF parts, despite the stochastic nature of the manufacturing process. Pores can come from the feedstock itself, from gas entrapment during the atomization process, or moisture on the surface of the particles. Porosity can also be induced through improper processing parameters. These parameters then result in a specific applied energy per unit volume, that is either too low, causing low fusion rate or poor powder consolidation, or too high, leading to overheating or keyholing. Pore morphology changes depending on the particular mechanism.

Internal flaws that form during processing, commonly known as lack of fusion defects, also negatively impact the fatigue behavior of components. In order to detect these internal pores, non-destructive x-ray computed tomography (XCT) is routinely employed, though it is limited by the resolution and analysis software. Sundar et. al. [144], published a study showing how XCT was only able to detect about 15% of the flaws compared to destructive serial sectioning. It highlights the need for better processing quality such that reliance on flaw detection through XCT is unnecessary.

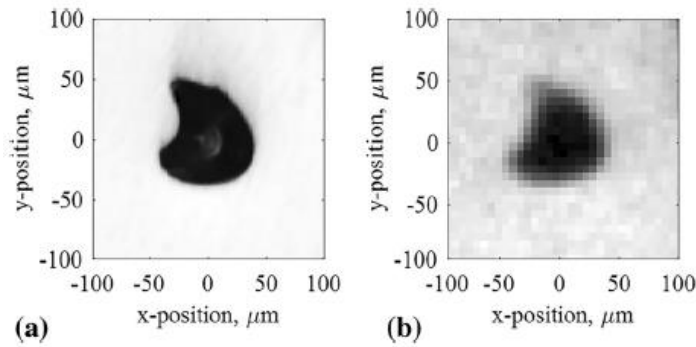


Figure 2.36. Lack of fusion flaw comparison between detection by a) serial sectioning, and b) XCT [144].

While the specific strength of Ti-6Al-4V is one of its most attractive properties, the alloy has also been shown to provide good performance under sliding wear conditions as characterized by microscratch, micropillar compression, nanoindentation, and microhardness measurements [145]. However, the as-built surface roughness, which plays into the material strength of PBF Ti-6Al-4V, is known to be difficult to characterize with traditional areal parameters such as S_a , S_z , or S_q [146]. Under cyclic loading, surface roughness can accelerate crack initiation causing early failure. Lee et. al. [147] proposed a hybrid surface parameter that consisted of the maximum valley depth (S_v), skewness (S_{sk}), and kurtosis (S_{ku}), in order to provide better correlation between the fatigue life and surface roughness.

Reddy et. al. [148], provided a framework to optimize the L-PBF process for minimizing defects in Ti-6Al-4V to maximize fatigue performance. From 6 mm x 6 mm x 73 mm 4-point bend printed fatigue bars, they machined 0.5 mm from each face to eliminate surface roughness caused by contour scanning. They were able to define a fatigue-based process window between beam power and speed (P-V) that consistently resulted in increased fatigue properties over regions outside that window. At a constant power of 370 W, beam velocities between 1050-1450 mm/s demonstrated higher fatigue strength, which corresponded to the speeds with the smallest, and lowest number of pores. That process window range is much narrower than a previously described window of 1150-1850 mm/s between generating keyhole defects or balling defects. Beyond 1500 mm/s, the irregular lack-of-fusion pores increased significantly in both quantity and size. A value of 1 pore/25 mm³ was defined as “d1,” and it correlated defect populations, crack initiation defect sizes, with highest fatigue properties.

Characterizing the surface topology of an additively manufactured component can depend on the measurement instrument. In a paper published by Thompson et. al. [149], the authors employed Confocal Microscopy (CM), Coherence Scanning Interferometry (CSI), Focus Variation (FV), and XCT to measure a sample surface, shown in Figure 2.37, produced by a Renishaw L-PBF machine. They computed the areal texture parameters (S_a , S_q , S_{sk} , S_{ku} , and S_{al} defined by ISO 25178-2) and discussed the discrepancies. Skewness and kurtosis parameters, when combined, confirmed the poor ability for optical technologies like FV to capture deep recesses, since it requires sufficient reflected light to compute contrast where XCT was suitable. The most capable methods for acquiring higher-frequency topographic content were CS and XCT, although some of which could be noise since the topographic features of interest can be as large as the measurement error.

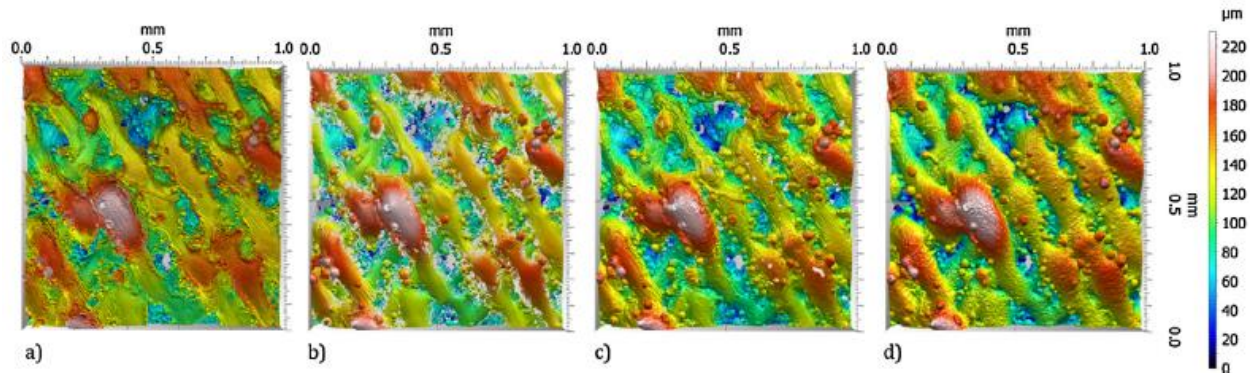


Figure 2.37. Height based coloring and top views of reconstructed topographies by a) CM, b) CSI, c) FV, and d) XCT [149].

In a study published by Lou et. al. [150], they investigated the application of the material ratio (MR) curve to characterize topological features such as open surface pores in AM parts. Originally named the Abbott-Firestone curve, and primarily used by the automotive industry for quality control of engine cylinder liners, it is a curve of the material-to-air ratio of a surface as a function of depth, and it is also described in ISO 25178-2. Three surface sets were measured with XCT and then analyzed via MR curves where typically anything below 10% volume (Mr1) was considered a peak, and above 80% a valley (Mr2). The MR curves proved capable of providing much richer characterization of a surface topology for comparisons than the limited S_a value method. It also provides another useful parameter known as the valley void volume (V_{vv}) that indicates the void volume per unit area for the characterization of open surface pores.

There is on-going research into other methods of characterizing the as-built surface roughness, such as random graph representation that captures the spatial distribution of morphological features [151]. The graph is characterized by the Fiedler value which is the algebraic connectivity that captures how well nodes of the graph are connected. It was studied to quantify the initial surface and the subsequent levels of surface finishing through post-process polishing. Another,

called Crossing-the-line segmentation [152], builds on ISO 4287:1997 for calculating feature parameters such as RS_m and R_c in a more robust manner to ensure feature-based evaluations are comparable. The algorithm detects hills and dales (peaks and valleys in ISO 4287:1997) defined by mathematical criteria, it applies a vertical threshold to eliminate insignificant features near the zero plane, and it merges smaller adjacent hills and dales. It was reported that the unambiguous implementation of this segmentation method was to be included in the profile standard ISO 21920.

The random, irregular additively manufactured surfaces require the use of statistical methods for predicting dimensional deviation. Characterization of the surfaces only describes the result following fabrication. That manufactured surface variability information must be taken into account by engineers when designing an additive assembly. The goal of tolerance analysis is to predict the final quality during the design stage with the main objective to estimate the probability of assemblability. The study by Dantan et. al. [153], presented mathematical models for the geometrical behavior prediction of an additively manufactured assembly, as well as an approach for tolerance analysis.

Shanbhag and Vlasea [154], looked into the effect of the EB PBF process on Ti-6Al-4V powder characteristics as the powder was reused over multiple cycles. They demonstrated how the morphology degrades due to the partial sintering and melting, as well as the blasting powder recovery process. As the PSD narrowed compared to virgin powder, the flowability and spreadability became worse. In a later powder reuse study [155], the same authors showed through SEM images in Figure 2.38, the extensive modifications and damage done to the surface of the particles with reuse. They then proposed a unified metric for scoring powder quality based on normalized values of measured morphology, size distribution, Hausner ratio and Carr index, basic

flow energy, specific energy, oxygen and nitrogen content, bulk density and tap density after 500 taps, cohesive index, and angle of repose.

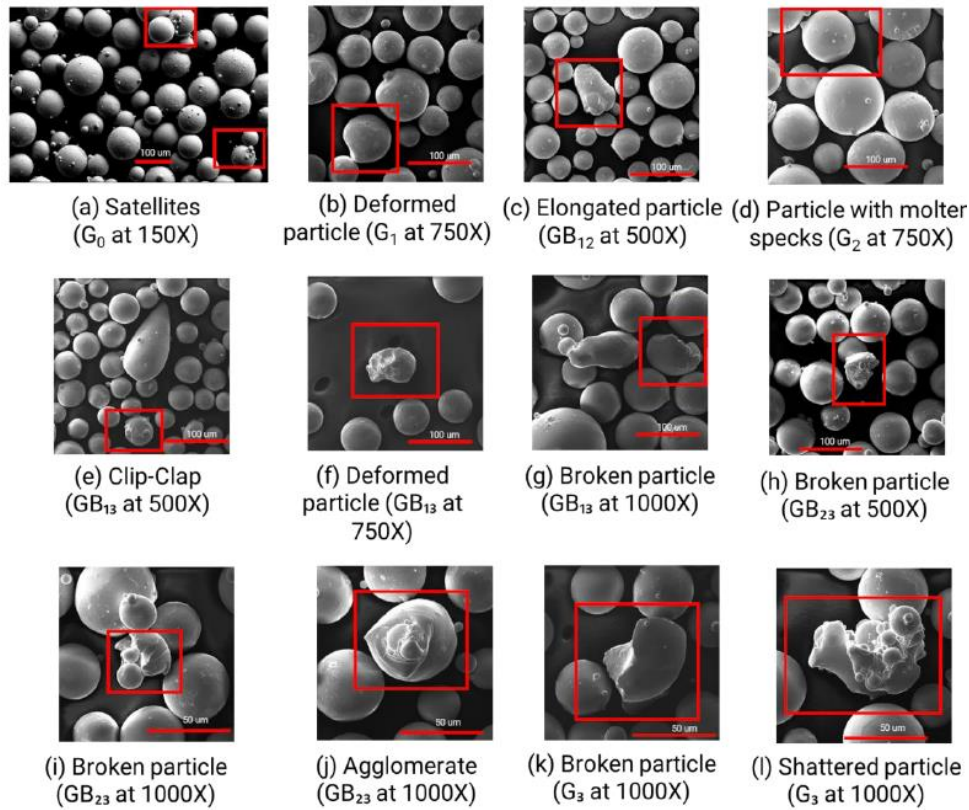


Figure 2.38. Examples of various powder types detected by SEM [155].

The effects of build plate location on the properties of Ti-6Al-4V, fabricated via EB-PBF, were evaluated by Gungor et. al. [156]. They compared ASTM E8 tensile specimens produced in each of the 4 corners and center of the build plate, all oriented vertically, in the as-built and post-processed machined state. In total, they had 50 specimens and employed a statistical analysis with and without spatial variability as a factor. Estimated probability density function model parameters were identified, followed by goodness-of-fit tests such as the Kolmogorov-Smirnov and Anderson-Darling tests, to select the best-fitted models. This ensured that the experimental results agreed well with the probability density function models, and that the number of samples

was sufficient. When spatial variability was included in the analysis, the authors implemented semivariogram plots. These plots illustrated that there is special variability in mechanical properties between specimens manufactured at the center of the build plate and those built in the corners. Machining the specimens was found to reduce the spatial variability, but did not eliminate it.

Because lattices like the Octet have struts produced in a variety of orientations relative to the build plane, it is important understand the effect build orientation has such that uniform lattice truss members will not have uniform mechanical performance. Tensile properties of Ti-6Al-4V lattice struts, manufactured via EB-PBF, were studied by Sepe et. al. [157]. In this case three strut diameters were studied (1, 1.5, 2 mm) at three build orientations (0, 45, 90 degree). The sections of their tensile coupons were printed using the Arcam A2X Net theme, as opposed to the Melt theme, which was specifically designed for lattices of 2 mm or less in diameter. Trusses with a 1 mm diameter showed inconsistent properties regardless of orientation as the performance was surface defect driven.

The authors [157], also found that those manufactured at 45 degrees out performed those at 90 degrees, which is consistent with expectations of the vertical (z-axis) orientation being weaker than the x-y orientation. However, defects in the horizontal orientation hurt the performance of struts in all diameters relative to the other orientations. Another factor that likely contributed to the low performance was the long span of the specimens that was not provided any supports in the build. It was interesting that none of the properties for yield stress, maximum tensile stress, or elastic modulus, were able to meet the expected properties of the parent material, and elongation was poor with a maximum of $< 4\%$ coming from the 90 degree 2 mm specimens. Fracture surface analysis displayed in Figure 2.39, was performed via SEM, showcased lack of fusion pores and

unmelted powder with the presence of a large number of fine dimples. The cross section of the 1 mm diameter specimens displayed a smaller and very irregular profile from the nominal circular section, and while still present to some degree, that aspect was less evident in larger diameter profiles.

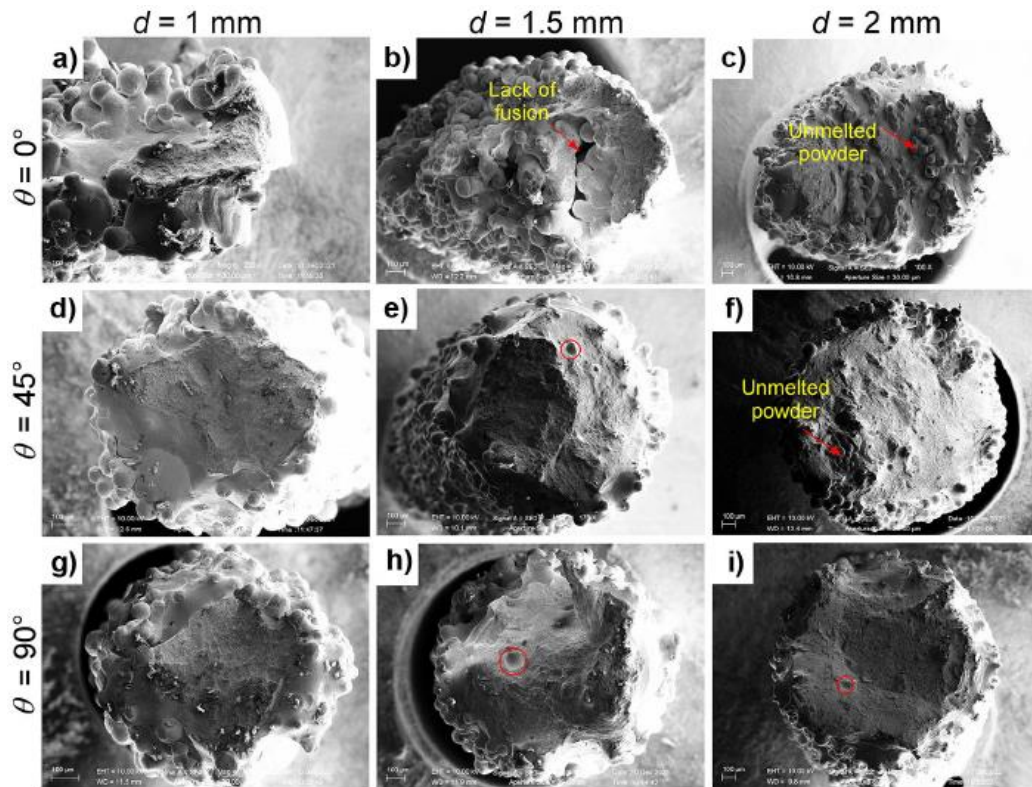


Figure 2.39. SEM images of fracture surfaces of specimens designed to mimic strut lattice geometry, built at different orientations and diameters [157].

Arrieta et. al. [158], used digital image correlation to also investigate the mechanical behavior of differently oriented Ti-6Al4V produced by EB-PBF. They performed ASTM E8 tensile tests at 0, 30, 60, and 90 degrees relative to the build plane, and then recorded the strain fields with digital imagery in the search for shear effects. While it was noted that the layering process produced uniform microstructures, and there were no obvious defects, the build angle may have

influenced the performance of the components. They reported the 30- and 60-degree specimens developed higher shear stresses that influenced the failure mode.

An investigation into thin-walled TPMS-sheet Gyroids produced by EB-PBF was conducted by Khrapov et. al. [159]. In an effort to reduce wall thickness, the authors assigned the Wafer theme, typically used for support material, for specimen fabrication. They varied the beam speed and current to determine optimal mechanical properties and geometric features. They designed a 15.7 mm^3 test sample model with zero thickness Gyroid webs, and produced five different configurations with measured relative densities ranging from 16-22%. The average thickness achieved was about 300-400 μm ; however, as shown in Figure 2.40, the webs of all samples contained numerous through-hole defects with their size corresponding to the beam energy.

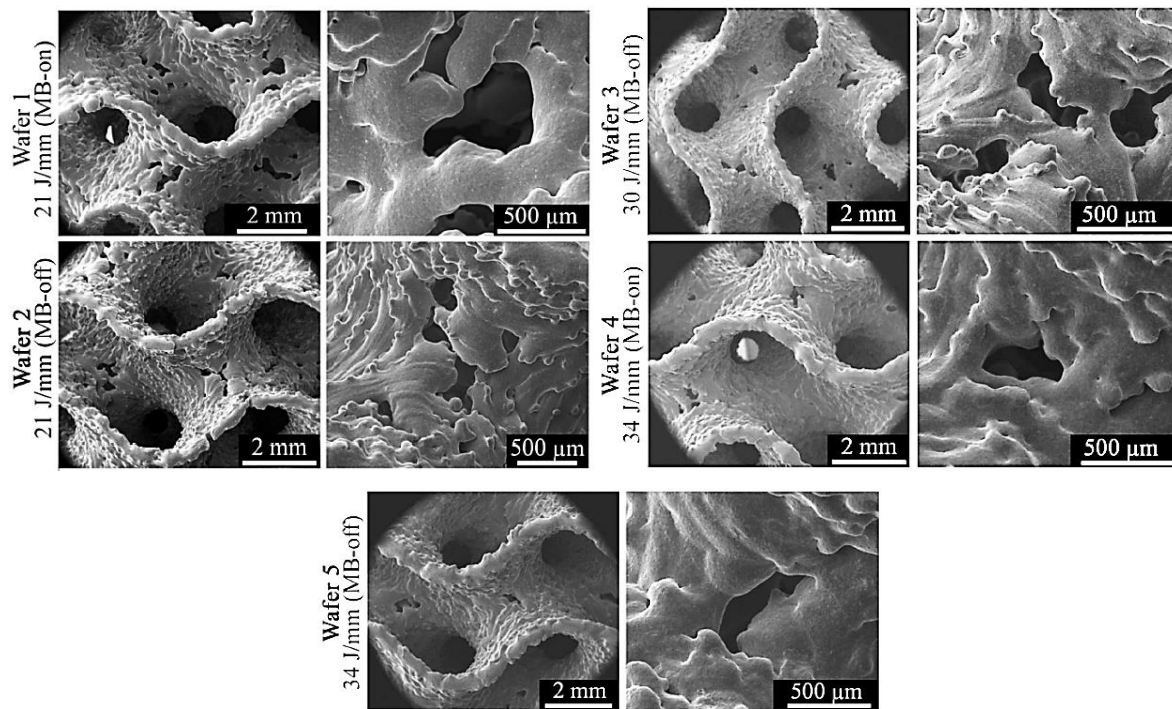


Figure 2.40. SEM images of EB-PBF Gyroid specimens manufactured with various settings at minimum thickness [159].

These specimens were tested in compression, and simulated in FEA with through hole defects, both showed that the defects did not have a significant effect on the quasi-elastic gradient with respect to relative density. Although, the defect size did considerably influence the mechanical behavior with the larger holes resulting in much larger variations. Microstructurally, they observed a lack of any β phase in the material, only α/α' martensite with laths in a Widmanstätten (basket-weave) pattern, which makes sense given the thickness and size of the specimens.

2.3.3 *Ti-6Al-4V Microstructure*

In EB-PBF, when producing fine lattice structures from Ti-6Al-4V, the melt process that occurs results in rapid cooling of the material, and it has been reported to form α' -martensitic microstructure (similar to what normally forms with L-PBF processes) [47], [48], [49], [51], [75]. This type of microstructure provides high static strength, but low ductility, often making AM lattices behave in a very brittle fashion during failure. Heat treatments have been shown to reduce or eliminate α' and convert it to a variety of microstructures that contain α and β phases depending on the cooling rate. This change in microstructure improves the ductility while decreasing overall strength [47], [48]. Therefore, a detailed review of microstructure formation research is necessary, with a focus towards Ti-6Al-4V produced via EB-PBF.

At the β transus temperature of 882.5 °C, titanium typically goes from a Hexagonal Close-Packed crystal structure (HCP, α phase) to a Body-Centered Cubic structure (BCC, β phase) in an allotropic phase transformation. The volume fraction of the α and β microstructure at room temperature is tailored with alloying elements, such as Al, C, N, or O, that are α -stabilizers, and Nb, Mo, Ni, Cu, and V which are β stabilizers. Conventionally produced Grade 5 Ti-6Al-4V alloy is considered a dual phase ($\alpha + \beta$)-type because of a slow cooling process [142]. The EB-PBF

process has higher beam energy than L-PBF, and operates at significantly elevated temperatures, where the cooling rate is also lower in a vacuum environment without the convective inert gas flow of L-PBF. Essentially, given sufficient build time, the EB-PBF process provides an in-situ aging heat treatment that results in little to no residual stress, and a unique microstructure.

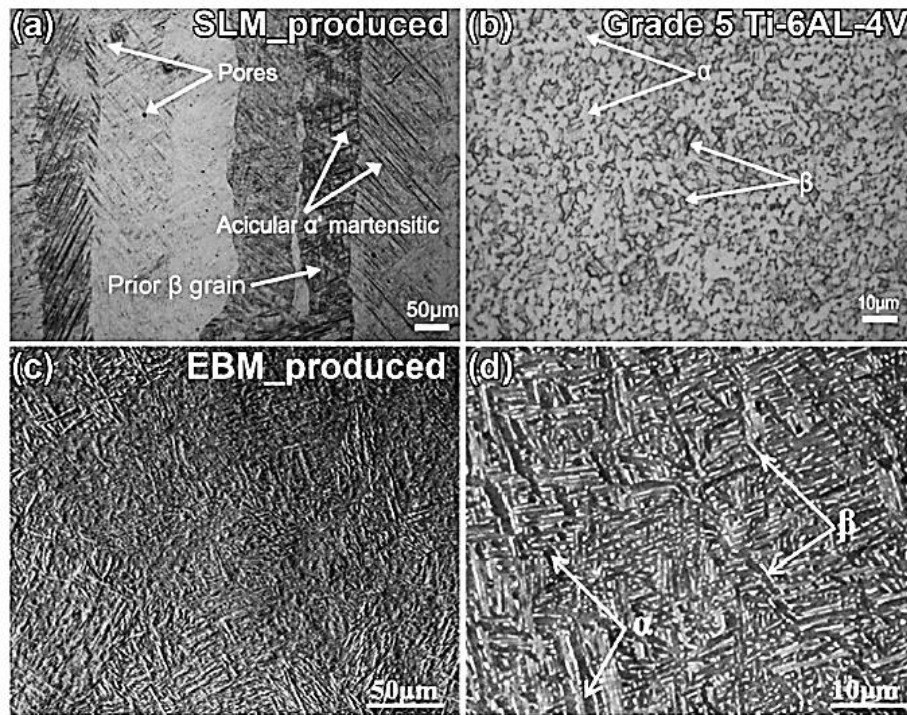


Figure 2.41. Microstructure images from a) L-PBF (i.e. SLM) produced Ti-6Al-4V, compared with Grade 5 Ti-6Al-4V bar stock, and c) and d) SEM images of EB-PBF (i.e. EBM) produced Ti-6Al-4V [142].

A large amount of α' martensite was found in Ti-6Al-4V produced via L-PBF because of the fast cooling rate, 10^3 - 10^8 K/s, relative to the critical cooling rate of 410 K/s for the β to α' martensitic transformation [142]. This forms ultrafine-grained acicular α' with β , that without heat treatment, trades enhanced tensile strength for the loss of ductility, causing the Ti-6Al-4V to be more brittle. The same would be true for EB-PBF if not for the aging environment (~ 500 - 700 °C in the build chamber for several hours), that decomposes α' martensite into fine lamellar α and

β . EB-PBF produced Ti-6Al-4V microstructure, shown in Figure 2.41, appears as a basket-weave (Widmanstätten) such that wavy columnar prior β grains are surrounded at the grain boundaries with lamellar α phase.

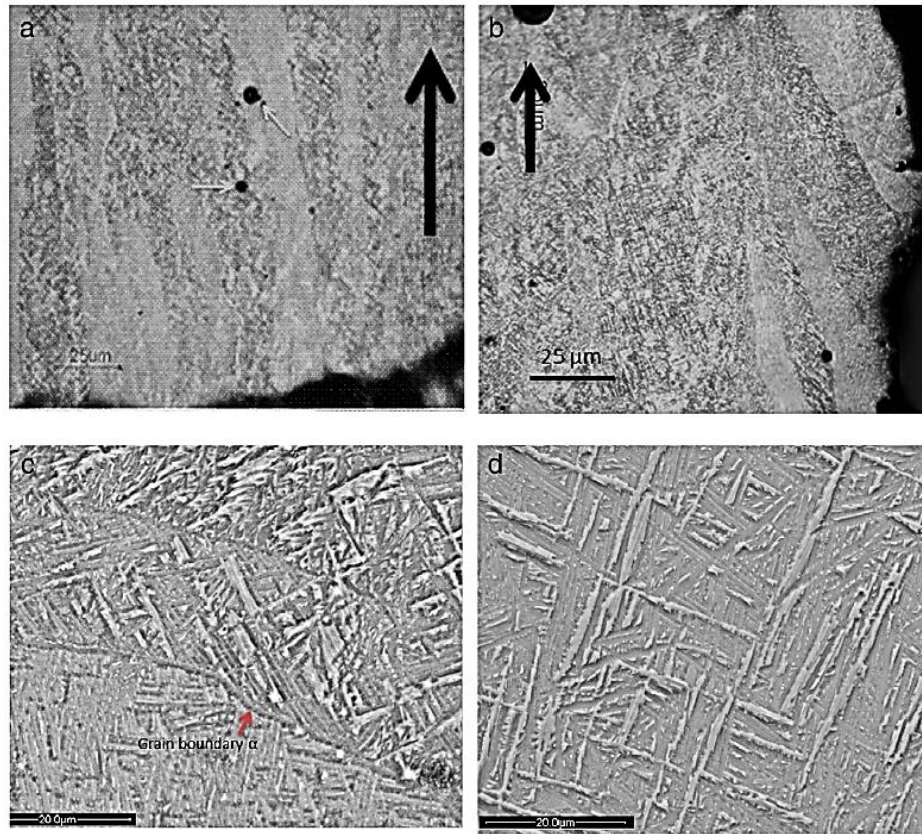


Figure 2.42. a) Optical micrograph of Ti-6Al-4V columnar microstructure with the build direction indicated, b) columnar microstructure near the surface tracking the temperature gradient, c) SEM of a typical columnar grain boundary, and d) α -platelets and α -colonies [160].

Microstructure of Ti-6Al-4V is processing history dependent, and is difficult to predict in EB-PBF due to the complex thermal history the material goes through. Thus, Safdar et. al. [160], sought to understand the Ti-6Al-4V microstructure at different stages of the build process, and produced tensile bars from gas atomized powder with a PSD of 25-40 μm . As displayed in Figure 2.42, they found that the microstructure was very fine compared to wrought, and that it contained

columnar grains of prior β growing in the build direction, with Widmanstätten α -platelets of different sizes and orientations. No distinctive layers were observed and the growth direction of the columnar grains deviated from the build direction near edges which seemed to track the temperature gradient. The diameter of the grains decreased with build height and ranged in diameter from 20 – 200 μm . Samples produced contained spherical gas pores (1-6 μm) that were non-uniformly distributed and attributed to the presence of Argon gas trapped in the powder from the atomization process.

Because PBF is similar to fusion welding at the micro scale, Ahluwalia et. al. [161], studied the $\beta \rightarrow \alpha + \beta$ transformation in titanium welds using a combination of a heat transfer FEM with a phase field solid-solid transformation model. Here, they used the FEM to compute the thermal history across the weld at the macroscale, in order to use it as an input to describe the microstructure evolution of the α phase precipitates in the β phase matrix, along with all possible variants, with the mesoscale phase field model. However, while this approach was not able to explicitly model the formation of α' martensite, the authors assumed martensite would form in a given location provided it did not undergo a change in composition by the time the martensite start temperature was reached during cooling. Using classical nucleation theory, they also assumed the nucleation to heterogeneous, and used experiments to assign a critical cooling rate of $\sim 2\text{ }^{\circ}\text{C s}^{-1}$ to avoid martensite formation.

The model showed that close to the weld line, the known rapid heating and cooling of the welding process, produced mixed microstructures of α , β , and α' , while further away non-equilibrium microstructures with partially dissolved α was observed [161]. These simulations help explain the commonly observed graded microstructures in titanium alloy heat affected zones

(HAZ) near a weld. The model identified processing conditions that would be useful in designing a process that would achieve a particular non-equilibrium microstructure.

When an EB-PBF build is carried out, phase transformation of the material from the liquified temperature of ~ 1900 °C to room temperature takes place in a three-step procedure [160]. First, the material experiences rapid cooling from ~ 1900 °C to ~ 700 °C at a cooling rate on the order of 10^3 - 10^5 K/s. Then the temperature is held at the build temperature of ~ 700 °C, followed by slow cooling to room temperature. During that time there are two potential phase transformation processes that could take place, either a diffusion-controlled transformation, or a transformation involving martensite formation.

A diffusion-controlled transformation would be where the temperature drops below the β transus, and the high temperature β phase (β_p) transforms into $\alpha + \beta$ microstructure during the isothermal hold at ~ 700 °C. During solidification, the prior β phase (β_p) is formed ($L \rightarrow \beta_p$), and then rapidly cooled to ~ 700 °C where $\beta_p \rightarrow \alpha_{GB}$, and Widmanstätten ($\alpha + \beta$) plates containing α grains and rod-shaped β phase are located on the α grain boundaries. The α phase is enriched in the Al stabilizer, while the rod-shaped β phase is enriched in the β species of Vanadium (V). This is similar to the isothermal transformation of wrought Ti-6Al-4V alloys and involves no martensitic transformation at any stage [160].

In the case of a transformation involving martensite formation, the prior β phase (β_p) would form during solidification ($L \rightarrow \beta_p$), and the Ti-6Al-4V rapidly cools from ~ 1900 °C to ~ 700 °C at a rate greater than 410 °C/s from single β phase to ($\alpha + \beta$), forming fully martensitic phase ($\beta_p \rightarrow \alpha'$). Subsequently during the temperature hold at ~ 700 °C, two possible transformations could occur: $\alpha' \rightarrow \alpha + \beta$, or $\alpha' \rightarrow \alpha'' + \beta$. The $\alpha' \rightarrow \alpha''$ transformation is thermodynamically unfavorable due to large lattice strain between the two phases, so α' formed during the initial rapid

cooling will transform to $\alpha + \beta$. The $\alpha' \rightarrow \alpha + \beta$ transformation is similar to conventional tempering at elevated temperature, where at 700 °C $\alpha' \rightarrow \alpha + \beta$ in about 30 minutes, which is much shorter than a typical EB-PBF build hold time. A small quantity of retained β phase may exist alongside α' , which is also transformed into $\alpha + \beta$ [160].

With that understanding, Safdar et. al. [160], reported that the diffusion-controlled reaction was the most likely transformation process based on their findings. They found that if α' formed during initial cooling in prior β columnar grains, then it would not be possible to form the α_{GB} layer around the grain boundaries. Also, reports of α' found in small 1-5 mm build samples was due to the lack of build time. It was noted that melting and solidification of each layer was not a single step phenomenon, because the prior β columnar grain height is much greater than the layer thickness. Regularity of the grain geometry depends on the cooling rate and the cooling path. Heat transfer to the previous layers acts as a source of continuous columnar grain growth along the build direction, but that near the part boundary, prior β phase columnar grains deviate from the build direction as some heat is lost to the surrounding powder [160].

A recent 2025 review of the primary nanophases in titanium alloys was published by Chen et. al. [162]. Throughout the processing of dual-phase titanium alloys, different process parameters, formation mechanisms, and the four types of nanophases (α , α' , α'' , and ω), can vary significantly, and affect the mechanical properties. They concluded that the nanoscale α and α' phase enhances strength (1200-1600 MPa), at the expense of ductility, and that the strength increases as the α size decreases. On the other hand, the α'' noticeably improved ductility and yield with lower strength (600-1300 MPa). Small quantities of ω phase had a similar effect as α on strength (800-1300 MPa) versus ductility, although excessive amounts led to brittleness. This understanding provides insights into how reinforced structures at the microscopic level can be achieved by finely tuning

element ratios, aging temperature and time, and controlling the size, morphology, and distribution of nanophases. Though, there is still insufficient knowledge when it comes to newer titanium alloys and processes.

The local temperature gradients during material solidification strongly affect the grain structure and texture, which influences a material's mechanical behavior. For this reason, Kergaßner et. al. [163], sought to model the grain growth during EB-PBF and use that as input to a meso-scale finite element mechanical model. From that multi-scale model, the authors were able to determine macroscopic elastic and plastic properties.

Popov et. al. [164], looked at the microstructural changes in Ti-6Al-4V specimens manufactured via EB-PBF with reused powder, and then also looked at microstructural changes after hot isostatic pressing (HIP). The phase recrystallization process that occurs during EB-PBF and HIP showed that the macrotexture of Ti-6Al-4V produced parts was preserved, similar to the crystallization texture formed during initial melting. Both microhardness and strength of the material increased because of oxygen dissolution in the α -phase. The authors reported that while powder reuse may be harmful for fatigue resistance, the static performance is not significantly affected.

Pegues et. al. [165], in part one of a series, focused on the relationship between the powder feedstock, manufacturing process, and post process conditions, and how the resulting microstructure and defects affected Ti-6Al-4V fatigue performance. They reported manufacturing induced lack-of-fusion defects and spherical pore aggregates produce a weak plane perpendicular to the build direction that results in anisotropic fatigue behavior. However, at the microstructural level they suggest that a functionally graded microstructure could be possible by actively

controlling the melt pool size and cooling rate. A broad range of phase structures is available such as: coarse $\alpha + \beta$, Widmanstätten (basket weave), and α' or α'' martensitic structures.

The effects of heat treatment on EB-PBF Ti-6Al-4V microstructure and properties were reported on by Galarraga et. al. [166]. Different heat treatments were performed and the effects of various cooling rates after solutionizing was studied. The solutionizing treatment was applied at a temperature 105 °C above β transus temperature (1100 °C) for 30 min, to ensure that the microstructure of the samples was 100% β phase. Then cooling was accomplished by furnace (~10,000 sec), air (~100 sec), and water (~2 sec), which caused the microstructure for the air and water to change from $\alpha + \beta$ to partial α' and full α' , respectively. The formation of hard α' martensite resulted in an increase in strength with a significant loss in ductility. The furnace-cooled equiaxed $\alpha + \beta$ showed decreased strength over the as-built columnar $\alpha + \beta$, and only a slight increase in percent elongation with more variability.

Additionally, the authors correlated variation of α lath thickness during aging with its subsequent effect on mechanical properties [166]. Experiments were conducted that combined solutionizing and aging to understand how they fit within the context of design and structural performance. Samples were solutionized at 925 °C for one hour to maintain the as-built columnar microstructure, and then air- and water-cooled before aged at 450 °C for 4 hours, followed by air cooling. Aging showed no effect on the microstructure of the water-cooled samples as the α' martensite did not decompose. The α lath thickness was shown to grow with temperature and time, which corresponds to differences in microstructures of parts close to the build platform that have been in the chamber longer than those near the top of the build. Aging time and temperature effect on α lath thickness ($\delta_{\alpha_{lath}}$ in μm) were related in Equation (2.83), with t in hours and the aging temperature T in K.

$$\delta_{\alpha_{lath}} = \delta_{\alpha_{lath}(t=0)}, \quad t^{\left(\frac{T-850}{1000}\right)} \quad (2.83)$$

Dislocation theory, where microstructural features function as dislocation slip obstacles, was assumed to explain the effect of α lath thickness on mechanical properties. The authors related Vickers micro-hardness (HV), yield strength (σ_{YS}), ultimate strength (σ_{UTS}), and elongation (el) using Hall-Petch type equations [166].

$$HV = 307 + 44 \delta_{\alpha_{lath}}^{-1/2} \quad (2.84)$$

$$\sigma_{YS} = 737 + 144 \delta_{\alpha_{lath}}^{-1/2} \quad (2.85)$$

$$\sigma_{UTS} = 986 - 5.4 \delta_{\alpha_{lath}} \quad (2.86)$$

$$el = 20 - 2.9 \delta_{\alpha_{lath}} \quad (2.87)$$

From their results, the aging temperature should not exceed 600 °C, because below that α lath growth is minimal, and the titanium will maintain desirable mechanical properties, ductility in particular.

Titanium is known to have strong corrosion resistance, often not requiring the priming and protection required by other common alloys in aerospace, like aluminum. However, there are still questions as to whether AM titanium alloys are more susceptible to corrosion than wrought products. A study conducted by Li et. al. [167], determined the influence of post process heat treatment and surface finishing on the corrosion behavior, and in the case of pitting corrosion, the $\alpha + \beta$ phase, with a finer grain α and higher β fraction, can provide high resistance. The different chemical composition in the α/β phases was shown to be susceptible to galvanic corrosion at the phase boundaries, as shown in Figure 2.43. The α -phase enriched Aluminum was dissolved as the anode by the β -phase enriched Vanadium that served as the cathode. The result showing that the

alloyed aluminum played an important role in reducing the corrosion resistance was not surprising given the well know susceptibility of aluminum to corrode.

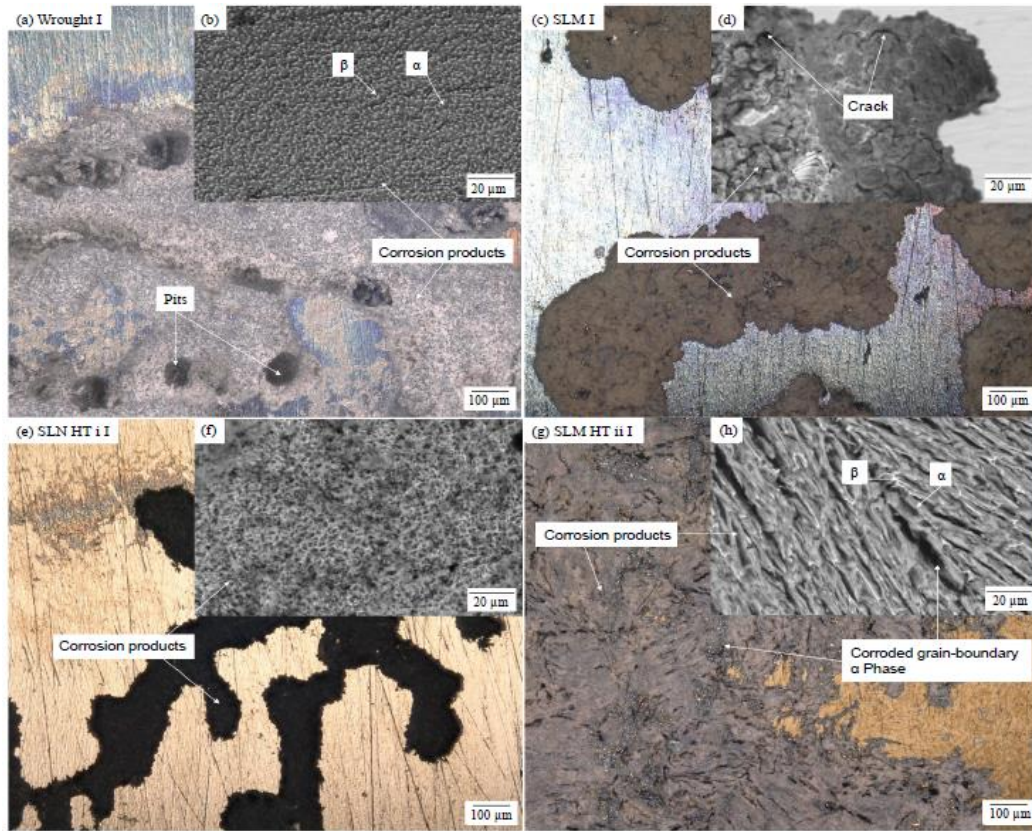


Figure 2.43. Examples of Ti-6Al-4V corrosion of wrought, L-PBF as-built, L-PBF heat treated (800 °C), and L-PBF heat treated (1200 °C) [167].

Titanium is also known to be a getter material such that it rapidly absorbs oxygen at elevated temperatures creating a brittle α' surface layer. The effect of oxygen pickup during L-PBF of a Ti-6Al-4V diamond-like lattice structure was studied by Velasco-Castro et. al. [168]. There, they purposefully produced increasingly brittle structures by increasing the number of scan passes from 1, to 3, to 5, and to 9, which were compressively tested. The higher number of passes created more heat and thus increased the relative density of the structures from 10.2%, to 14.9%, 18.4%, and 24.6%. Their experimental measurements did show increased oxygen pickup with the increase in

number of passes. They attributed that to the increase in the brittleness of the structures. However, they did not fully consider the presence of α' martensite despite the XRD samples showing a high alpha phase presence with no β -titanium peaks. It is likely the excessive heat buildup from the higher number of passes formed almost exclusively an α' martensitic grain structure that was the major contributing factor in the compression results, rather than the oxygen pickup that increased from 0.412 to 1.15 wt%. Notably, the study did not specify the grade of Ti-6Al-4V, the source, or whether it had been reused, but at 0.412 wt% it is more than double the maximum for a typical AMS4911 Grade 5 Ti-6Al-4V product.

In a powder-feed DED study with Ti-6Al-4V, Yang et. al. [169], characterized the behavior of fresh and recycled powder materials with varying levels of oxygen, up to 9500 ppm, mixed with the typical argon atmosphere. They showed that the ductility of the Ti-6Al-4V Grade 23 ELI samples, produced under a 9 ppm oxygen environment, had a strain-to-failure of 18.6%, while with 3500 ppm, the average elongation was reduced to 10.5%. Ductile fracture surfaces with fine dimple features turned into brittle features with inclusions and facets as oxygen exposure increased. Microstructurally, it was observed that the α lath thickness increased with more exposure to oxygen resulting in a larger presence of martensite. This is consistent with the increase in yield strength and ultimate tensile strength, which displayed a linear relationship with equivalent oxygen exposure.

Bruno, Rochman, and Cassar [170], reported on how build orientation influenced the microstructure and mechanical properties in EB-PBF Ti-6Al-4V, and compared their results to conventionally produced wrought Ti-6Al-4V. In the EB-PBF samples they observed columnar prior β grains following the build direction with a α lamellar/acicular microstructure within, whereas the wrought parts exhibited a more equiaxed microstructure. The microstructural

differences, in addition to defects, resulted in the wrought samples demonstrating a higher tensile capability than those produced by EB-PBF. Horizontally built specimens allowed for the longest scan lengths, and therefore the highest cooling rates, such that the resulting refined microstructure provided increased strength and lower ductility over samples built vertically, but that also increased the probability of defects that caused a large variation in the values reported.

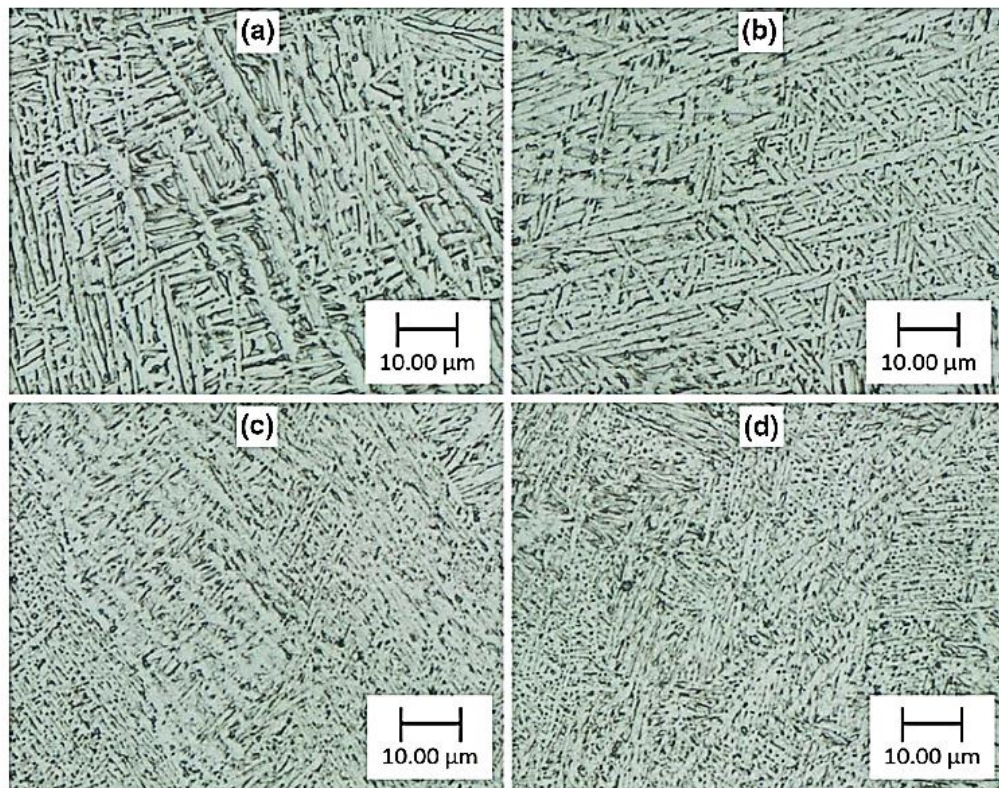


Figure 2.44. Ti-6Al-4V microstructure images from varying EB-PBF scan lengths: a) 30 mm, b) 50 mm, c) 70 mm, and d) 90 mm [171].

The effect of scan length on the mechanical properties and structure of EB-PBF produced Ti-6Al-4V was investigated by Everhart, Dinardo, and Barr [171]. They describe how the beam power and speed vary within the build, dependent upon the scan length, i.e. the distance the beam will travel in a particular direction during the hatching phase, which also depends upon the part geometry. They observed a dramatic drop off in part density at scan lengths of 80 mm or longer,

indicating process failure. At that point, the long scan length resulted in excessive heat dissipation and lower residual heat from prior beam passes, such that more energy is needed to achieve full melting and wetting. As shown in Figure 2.44, changes in the microstructure with respect to scan length were also observed, whereas scan length increased from 30 mm to 90 mm, the average lath spacing decreased from 0.431 μm to 0.385 μm . This refinement in microstructure correlated with higher cooling rates during solidification, resulting in higher yield and ultimate strengths.

The columnar microstructures aligned with the build direction have been documented with other PBF produced materials. One study in particular, published by Maconachie et al. [172], analyzed the quasi-static and dynamic responses of L-PBF AlSi10Mg relative to the build direction. Through optical micrographs and SEM imaging, they revealed how the microstructure followed contours of overlapping elliptical melt pools based on the scanning direction. From tensile experiments, they found high correlation ($R^2 > 90\%$) with the Johnson-Cook strain hardening power law model shown in Equation (2.88) [173],

$$\sigma = A + B\varepsilon^n \quad (2.88)$$

where A is the yield strength of the material, and B and n describe the strain hardening. Fracture plane angles were seen to vary, depending on build direction, based on optical analysis of the fracture surfaces that showed it occurred along melt pool boundaries. While the authors did observe a difference in ductility relative to the build direction, with those built parallel to the build direction exhibiting less strain to failure, they were not able to identify any significant difference in the hardening behavior due to build orientation or strain rate.

Ladani reported on the effects of build orientation, and relative distance from the build plate, using Ti-6Al-4V in EB-PBF [174]. She examined the mechanical properties and microstructure of ASME E8 M-04 dog-bone specimens placed in the vertical and horizontal orientations relative

to the build direction. It was noted that the vertical specimens showed significantly lower strength than the horizontal specimens, but did see higher ductility. Strain rate sensitivity (SRS) was determined based on Equations (2.89) and (2.90),

$$m_{\text{uniaxial}} = \frac{d(\ln(\sigma))}{d(\ln(\dot{\epsilon}))} \quad (2.89)$$

$$\sigma = k(\dot{\epsilon})^m \quad (2.90)$$

where m is the SRS exponent and was found to be 0.0374 for the vertical specimens. A positive SRS means higher strength at higher strain rates, and is typical of traditional Ti-6Al-4V, but the horizontal specimens did not demonstrate any particular SRS behavior.

In the microstructure analysis of the horizontal specimens, columnar prior β grains were seen growing in the vertical direction, parallel to the build direction, and containing α colonies that start to form at the grain boundaries and move inwards at an angle into the β grains [174]. However, in the vertical samples α and β phases were observed to be uniformly dispersed with no apparent columnar grains. This build direction columnar grain growth was described as the reason for the anisotropic elastic and plastic behavior. For the vertically built specimens, where loading was along the columnar prior β grain direction, slippage at the grain boundaries allowed for an earlier onset of yielding, whereas when loading was perpendicular to the columnar direction, deformation was limited by the shearing of α platelets and stopping at the β grain boundaries.

While most AM investigations have used long established alloys that were developed for other manufacturing processes, it is theorized amongst many researchers that the optimum alloy for PBF has yet to be discovered. An example of new alloy development and design opportunities using PBF was published by Zhang et. al. [175]. They produced an in-situ alloy (25Ti-0.25O) by mixing pre-alloyed Ti-6Al-4V powder, and CP-Ti powder, that were both doped with Fe_2O_3 nanoparticles via surface engineering, and then melting the feedstock within a L-PBF machine. The result was

an alloy that demonstrated superior mechanical properties to typical PBF fabricated Ti-6Al-4V, because it diluted the V concentration which eliminated phase heterogeneity issues, as well as compensated for any strength loss through the introduction of Fe and oxygen solute. In addition, spatially uniform lamellar ($\alpha + \beta$) microstructures were also reported in both the horizontal and vertical directions, avoiding the typical columnar grain growth in the build direction.

2.3.4 *Quality in Additive Manufacturing*

AM is widely recognized in its ability to produce parts with complex features; although, deviations in quality do exist, and they originate from a stakeholder's perspective of "fit, form, and function." For example, the as-built surface roughness might demand post-machining for any mating, bolting, or joining (i.e., fit) assembly operations, which also affects mechanical performance (i.e., function). Component failures typically originate at part surfaces where roughness and irregularities generate stress concentrations; this effect magnifies applied stresses and decreases the fracture and fatigue performance [40], [176], [177], [178]. The presence of defects within the bulk material exacerbates this decrease in function by serving as additional sites for fracture nucleation [35], [170], [171], [179], [180]. Since parts are constructed at elevated temperatures, anisotropy in the microstructure and material properties is a concern as it subsequently effects the function. Large thermal gradients that occur during cooling result in deviations from the nominal geometry (i.e., form) due to thermal strain mismatch.

As Uhlmann et. al. reported [181], more knowledge of PBF process technologies, and the resultant material properties, is essential to achieving upcoming fuel consumption and emissions requirements for the aerospace industry. In cooperation with Embraer and Fraunhofer IPK, the authors investigated the characteristics of titanium parts produced by L-PBF for aerospace structures. Through a wide range of techniques, they analyzed the material at each step of the

process flow from the raw material powder feedstock, to part production, all the way through post-processing and finishing steps. Their work demonstrated the various metallurgical changes that occurred throughout the different processes, and it highlighted the need for improvements in production and quality control before broad adoption by the aerospace industry could be obtained.

An experimental study by Gong et al. [182], demonstrated how the characteristics of energy input for PBF processes affected the origination of defects. It included detailed image analysis of the topological melt tracks at various beam powers with Ti-6Al-4V. High energy densities in the hatch lines caused overmelting, and defects generally resulted from vaporization within the melt pool. However, within the EB-PBF machines the adaptive beam control system makes it difficult for the creation of those types of defects. A phenomenon known as mass transfer can occur during the melt phase, and it can generate large particles welded to the upper surface at each layer that disrupt the powder distribution on the following layer. Melt pool discontinuities and insufficient melt pool overlap along hatch lines can also create defects between layers that result in porosity and lack of fusion [183], [184].

Understanding how fabrication risks affect part and process selection, control, and certification to stakeholder requirements remains an important challenge. It must be addressed if AM processes are to be applied to mission-critical functions where performance optimization, quality, and reliability are critical. The greatest barriers to wide-spread AM implementation and adoption have to do with repeatability and reproducibility limitations, which includes precision and dimensional accuracy [185]. Each design, build, and post-process parameter for production must be well-documented for quality assurance. Materials must also be of high-quality in order to meet demanding certification standards that vary depending on the applicable industry [11]. Parts must be inspected and tested in accordance with specifications that reflect stakeholder

requirements. Fabrication equipment, such as the AM system, must be monitored and maintained with controls for documenting changes.

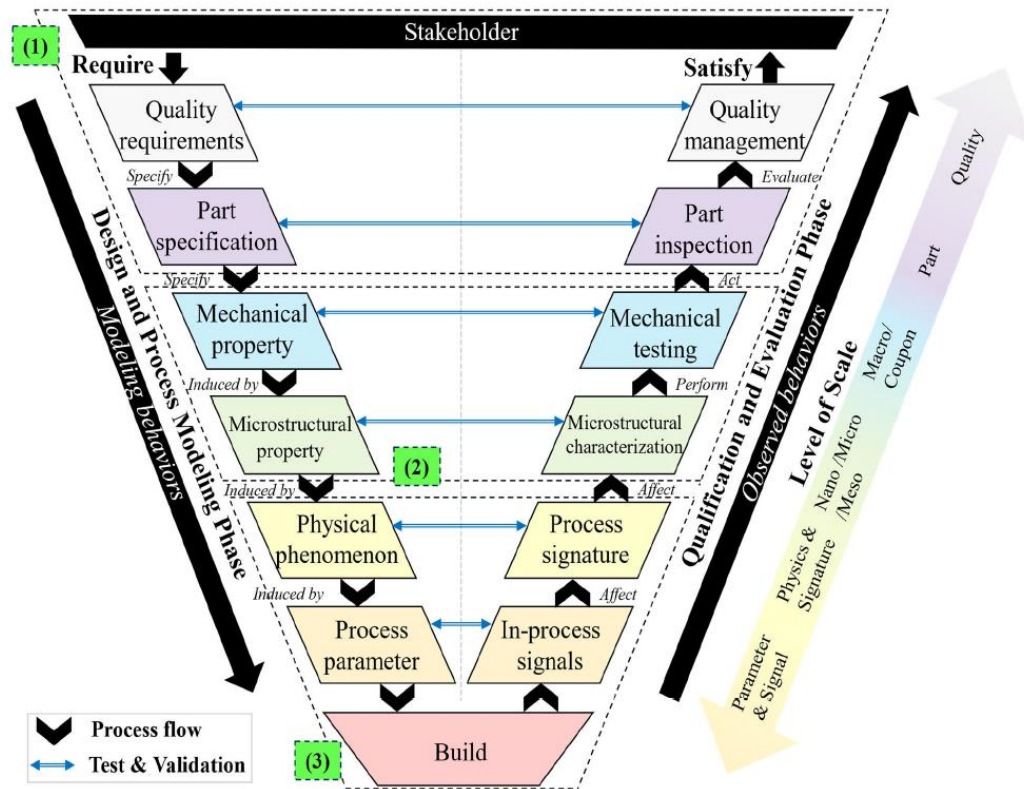


Figure 2.45. A systems engineering V-model with respect to quality assurance in metal AM [186].

A V-model framework for ensuring quality in metal AM components was proposed by Roh et. al. [186], and displayed in Figure 2.45. Their V-model was adapted from the systems engineering “Systems V” popularized by NASA [187] as a way of visually depicting the process of validation and verification. This representation helps ensure all stakeholders are involved from the development of initial quality requirements, to the final management and maintenance phase, and it employs verification at every step of the process. The left side of the V-model begins with the specification of quality requirements, which flow down into the part specification, mechanical properties, microstructural properties, physical phenomenon, process parameters, and ends with

the AM build. Those steps are matched in reverse order by the right side flowing up from the build with in-process signals, process signatures, microstructural characterization, mechanical testing, part inspection, and finally quality management. This V-model framework for metal AM serves as a good way to facilitate part qualification through quality assurance and process maintenance.

A study conducted by Kelley and Ramulu [188], investigated the EB-PBF process-to-part quality relationship in terms of external and internal defects, and component microstructure relative to a feedstock that had been reused for 33 build cycles. Using optical microscopy, they characterized the particle morphology using shape factors, and presented a method to define low frequency, macro-scale particle shapes implicitly using a binarized domain. Correlations were found between certain classifications of irregular particles and lack of fusion defect formation. A more noticeable effect on surface roughness and microstructure formation was observed based on beam power and scan speed parameters, rather than local particle morphology. This work showed that, with a high enough line energy density, irregular particles could be used to create defect-free parts, assuming any challenges in powder distribution were negated.

Gaikwad et. al. [189], used in-situ monitoring of the melt pool dynamics with a pyrometer and high-speed video to capture the quality of single melt tracks in L-PBF. Data was collected over a wide range of laser power (P) and velocity (V) settings, and then processed by a Sequential Decision Analysis Neural Network (SeDANN) model in order to develop predictive quality capabilities. They analyzed height map measurements of the single-track melt morphology for the width and percent continuity. Quality labels such as keyholing, conduction, lack-of-fusion, and balling, were then generated for the various morphologies based the linear energy density $E_L = P/V$. While power and velocity are not the only important process parameters to good melt

quality, the neural network method employed in that study showed that it could be expanded to incorporate more parameters like hatch spacing and spot size, as well as additional sensors.

Novel or improved flaw detection inspection methods are being developed to inspect additive parts, particularly those with internal complex lattices for lightweighting. Obaton et al [190], experimented with volumetric nondestructive testing on samples using resonant ultrasonic spectroscopy, impulse excitation technique, and electromagnetic-acoustic resonance. While these various methods had reasonably good success in detecting a good part from one with internal flaws, the size and nature of the flaws could not be resolved. Thus, the need for more research and new methods development was advocated.

A study was performed by Downing et. al. [191], that attempted to use a μ CT scan to create a virtual stylus with which to measure the surface roughness of Ti-6Al-4V L-PBF specimens. While CT does allow for non-destructive imaging of internal surfaces, roughness is not directly observable from the data. The goal of their method was to enable the extraction of traditional profile parameters, and they used solid cylinders as samples such that the experimental results would be relevant to AM lattice structures. Because CT provides a stack of cross section images where surface profiles could be extracted, sufficient length had to be determined for each roughness parameter (R_a , R_q , R_p , R_v , and R_z) for its finite integral to converge on the limit value within a specified tolerance.

Equation (2.91) describes a proxy surface profile $z(x)$ as a sinusoidal waveform where the profile amplitude (a) and the wavelength (λ) of the sine wave were used to determine the sample length (l) by evaluating the integrals as $l \rightarrow \infty$ [191].

$$z(x) = a \sin\left(\frac{2\pi x}{\lambda}\right) \quad (2.91)$$

Evaluating convergence results indicated the theoretical minimum sample lengths for each roughness parameter, which was then used to determine threshold values for adequate resolution of features from the image voxels. Voxelized representations of surface geometry still leads to error in measurement accuracy, yet the authors found that a voxel size of $1/5^{\text{th}}$ of the mean peak lateral spacing was sufficient to keep roughness parameters within 10% of measurements performed with optical microscopy. However, it was noted that the minimum voxel size is at the limits of resolution of the existing CT detectors. and that there is a need to improve the minimum voxel size to bring it closer to the smallest measurable roughness. The benefit though is in the ability to measure the roughness of internal surfaces, like those of strut and TPMS lattices cores in sandwich structures, that could make this an important approach to consider.

One method for the non-destructive evaluation (NDE) of AM parts was demonstrated using an ultrasonic C-scan [183]. In that work, artificial flaws were placed within L-PBF Ti-6Al-4V samples in order to assess the ability of the NDE method to detect flaws that ranged from 200-500 μm in size. While the analysis did reveal the known flaws, the precision was low, such that it was unclear whether the detection method would successfully identify defects that were not known, or be able to adequately characterize them at that scale. The results indicated more of a binary output, i.e. is there something there or not, which may then drive a more detailed evaluation such as μCT .

Current quantification methods for reporting porosity in AM parts does not provide information about pore size, shape and distribution, all of which can have significant impact on quality and mechanical reliability. Varney et. al. [192], demonstrated a method of characterizing pores in EB-PBF Ti-6Al-4V by using knowledge about the formation mechanisms combined with fatigue testing and microstructural characterization. They demonstrated how percent porosity cannot solely determine mechanical properties because localized effects led to variations as high

as 10 times the average value. Circularity was identified as a key parameter with 0.8 dividing pores into two groups, spherical and lenticular, both with distinct formation mechanisms. Despite a constant pore volume, as the shape becomes more lenticular, which tends to be perpendicular to the build direction, the estimated fatigue life can be reduced by almost 10% according to the Murakami model.

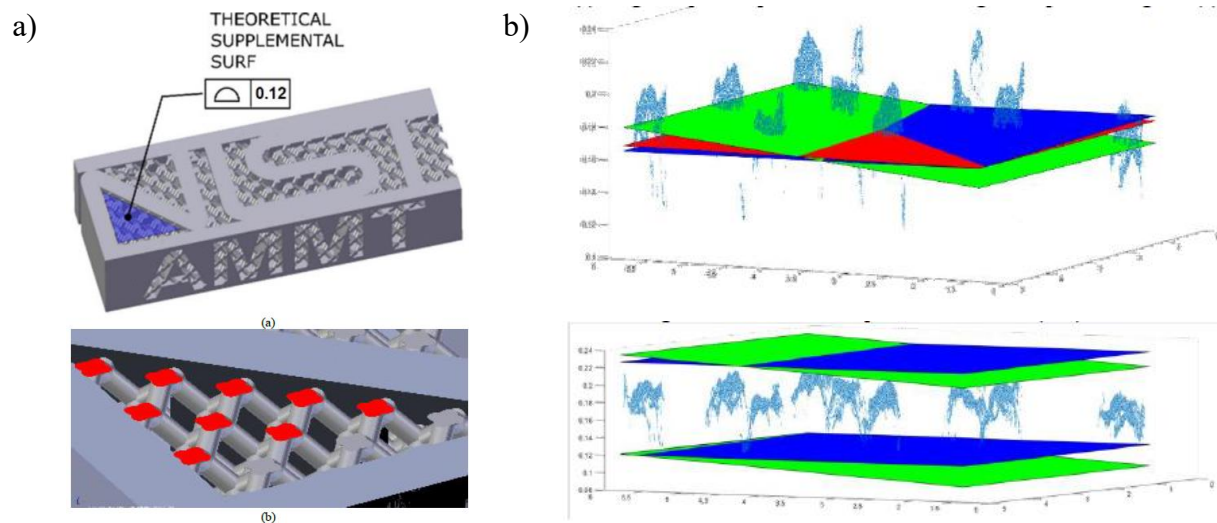


Figure 2.46. a) Quality definition of a Lattice structure theoretical supplemental surface (TSS), and b) comparison of Least Squares (green) and Chebyshev (blue) method for inspection of the TSS (red) flatness [193].

For light-weighting applications, where lattice structures may be applied as infill replacement for solid material, the lattice must follow a supplemental curve, surface, or volume, that is inspectable to a part specification. Ameta, Fox, and Witherell [193], employed an artifact developed by the National Institute of Standards and Technology (NIST) to conduct a case study for tolerancing a Theoretical Supplemental Surface (TSS) inferred using geometric elements from the lattice, to a Derived Supplemental Surface (DSS), the laser scanned profile derived from the physical part. They used two different fitting algorithms, Least Squares and Chebyshev, to

determine how well the lattice DSS conformed to the TSS. The study highlighted the difficulty in inspecting lattices to a part specification, because as shown in Figure 2.46, the Chebyshev method found the part was in conformance while the Least Squares method did not.

Dallago et al [194], inspected the thickness of struts in uniform cubic lattice specimens produced via L-PBF. The lattice unit cell size was 4 mm with circular struts of diameter 0.760 mm, and the inspection methods employed were 3D point clouds from CT reconstructions, metallographic specimens, and optical light microscopy. They observed the elliptical nature of horizontal struts due to the poor down-skin surface roughness, while the vertical struts aligned with the build direction were consistently circular. Their conclusions were that optical microscopy was most susceptible to surface roughness irregularities that resulted in higher values of strut cross section parameters by 5% for vertical struts and 10%-20% for horizontal struts. The destructive metallographic technique, while comparable in accuracy to the other methods, provided too few data points to support a statistical analysis and specimen preparation was labor intensive. CT and optical approaches were both effective at capturing the larger bulk volume based on limited small samples; however, CT scans were much more expensive and time consuming to obtain.

A study of shape and dimensional accuracy of Ti-6Al-4V lattices produced via L-PBF was conducted by Alkentar and Mankovits [195]. They evaluated three different strut type lattices produced by the ANSYS SpaceClaim application, with strut thicknesses set to 0.7 mm, unit cell lengths ranging from 1.8 mm to 3.8 mm, and relative densities from 26% to 31.6%. This is a rather new method of generating a lattice, where two of the structures (Double Pyramid Lattice with Face Diagonals, and Octahedral Lattice 2) appeared non-uniform and stochastic versus the third which was only identified as 3D Lattice Infill Pattern. The octahedral lattice resembled closely a traditional Octet lattice, except that the geometry was not periodic.

Another aspect that was new was the GOM ATOS high-speed laser scanning system they employed to graphically compare the as-built lattices to the CAD models [195]. There were no theoretical aspects aside from basic calculations of volume fraction and density. The laser scans revealed excess material agglomerated around nodes and along downskin surfaces which is to be expected in unsupported additively manufactured features. Dimensionally, they related the complexity of the structure to the level of deviation, qualitatively stating that the more complex the structure, the higher the level of deviation. Thus, concluding with the recommendation that a simple structure would be more accurately produced; however, the terms ‘simple’ or ‘complex’ are subjective and only based on the three particular lattice structures examined in this study.

Almalki et. al. [196], looked into cross-section shapes of lattice struts and the resulting shape properties. Their analysis method involved calculating the second area moment, radius of gyration, elastic and failure shape factors for bending, and the effective diameter. They used a L-PBF machine to print long extrusions of circles, octagons, triangles, and squares, at 35, 45, and 90 degrees build orientation, in four different nominal diameters of 0.5, 1, 2, and 3 mm, out of Ti-6Al-4V, and AlSi10Mg. The experimental parts were scanned via μ CT and then compared to the idealized CAD sections. A DOE main effects analysis shown in Figure 2.47 revealed that effective diameter has a major impact on lattice element manufacturability, particularly in titanium as the strut size increased from 0.5 to 3.0 mm. Also with effective diameter increasing, the shape factor measurements more closely aligned with nominal values meaning greater print accuracy, especially important with titanium. These effects demonstrated how printing minimal thickness lattices resulted in lower structural performance.

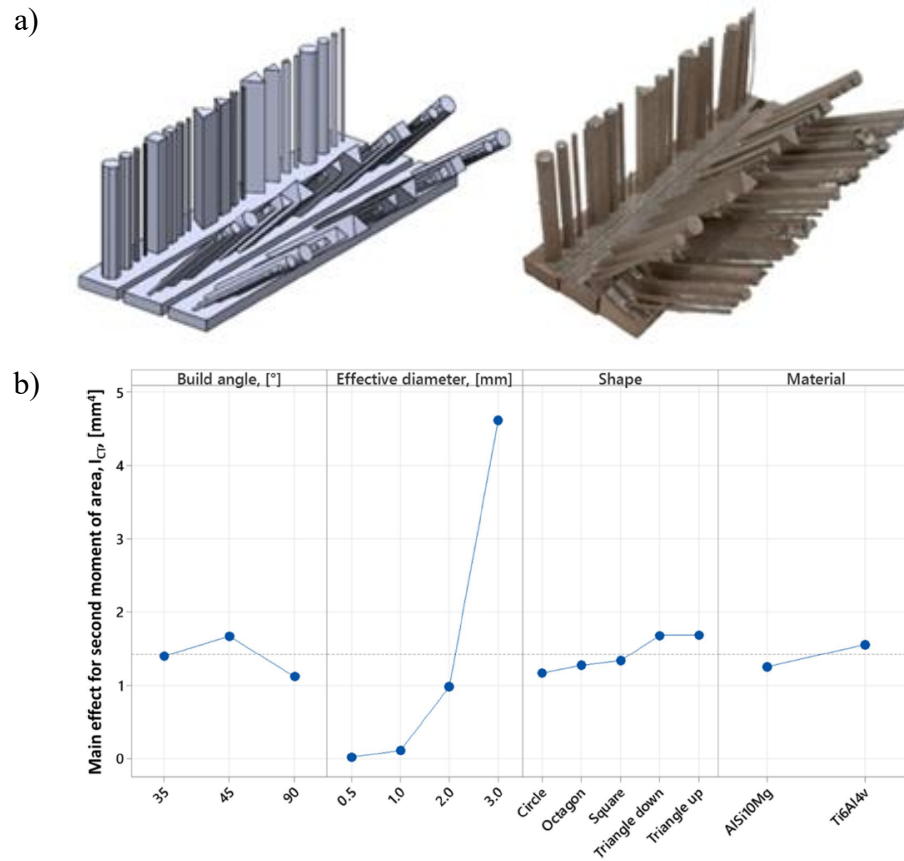


Figure 2.47. a) Design of Experiments with lattice strut geometries CAD model and as-printed, and b) Second moment of area main effect plot among independent variables [196].

It becomes apparent that new manufacturing processes are nearing maturity when research begins emerging with a focus on quality assurance while pivoting away from whether it can be done. This is the case with PBF processes, as more attention is starting to be paid to manufacturing quality and repeatability. As the technology turns this critical corner, any future researchers that do not at least address the quality of the experimental specimens is not contributing much to the state-of-the-art. While existing methods of inspection have provided some insight, research so far has shown that additional methods and techniques are required to meet the demands of increased geometric complexity.

2.4 SUMMARY

Amongst the literature surveyed, the use of EB-PBF has not been as extensive as L-PBF, despite the many process advantages such as speed, beam energy density, elevated build temperature, and the vacuum environment. Powder bed electrostatic charging is one of those unique concerns for EB-PBF; however, research has shown that the critical charge level can be mitigated through a variety of methods to tune the process parameters. The commercial approach has been to utilize a feedstock powder with a larger PSD, since gravity has shown to be the strongest influence in resisting powder motion, as well as maintaining high powder bed temperatures. Although, if an alternative feedstock is utilized, reduced beam current and high beam speeds during the preheating phase will help to suppress the smoking phenomenon.

Powder bed fusion has been an enabler for manufacturing architected lattice structures, which until recently have been impossible to produce. Simple strut type lattices have been the most common in literature because lattices were also challenging to design using explicit CAD modeling tools. Implicit modeling tools, such as MATLAB and nTopology, have opened the design space to include more lattice types like TPMS structures that have smooth surfaces and contours which lend themselves to be self-supporting in AM processes. Topology optimization methods have also begun to be demonstrated within the architected materials space with functionally graded topologies and morphologies. When integrated into a metallic sandwich structure, the customization afforded the designer to tailor the material system to the application is unmatched by any other manufacturing process.

To this date, much of the focus has been on characterizing the properties of micro lattices as a bulk material in compression based on the relative density. However, one of the biggest problems with this approach is that the load carrying member size is on the same length scale as surface and

manufacturing defects. If the quality of the lattice structure or the fabricated material is poor, then the defects will dominate the material behavior rather than the design of the structure. Therefore, there is a lot of room still to develop architected lattices with sufficient quality as to determine differences in geometry. There is very little research into the flexural behavior of these novel topologies under more realistic loading conditions, which requires incorporating the architected lattice into a sandwich structure, and demonstrate the performance empirically while following guidance from industry standards.

Ti-6Al-4V components produced via EB-PBF have been shown to likely have fewer internal defects and a more favorable $\alpha + \beta$ Widmanstätten microstructure than L-PBF, without post-processing. While mechanical properties of Ti-6Al-4V have been long understood, and existing constitutive models may work well in some cases, they make many assumptions or have parameters that must be derived empirically. When dealing with anisotropic printed materials or topology optimized lattice structures, the popular method of homogenization becomes unusable, and more robust analytical models are required to address the future of mechanical structures. Quality assurance in PBF manufacturing is also an area that requires further research, but is only beginning to emerge. The NASA popularized Systems-V is a logical QA approach, although the practical implementation and application at all levels has not been sufficiently demonstrated throughout the available literature.

Based on the surveyed literature, there is a clear need for more work on characterizing the art-of-the-possible with topology optimized lattice sandwich structures. Additive manufacturing with Ti-6Al-4V by EB-PBF provides the capability to fabricate these complex designs that cannot be made any other way. In industries such as aerospace and medical, highly tailored and weight optimized structures are desired, because these structures affect the health and environmental

performance of their applications. While the medical industry focuses on the micro scale, and aerospace on the macro scale, the two play an important role in changing the manufacturing paradigm. Where conventional manufacturing techniques would first create the material, then the part, AM provides the ability to create the material and component at the same time. Exemplars will be important for changing perspectives in order to introduce a new generation of engineers to DfAM techniques and best practices.

Chapter 3. MOTIVATION, GOALS, AND OBJECTIVES

Throughout history, from a design perspective, new materials and manufacturing processes have been enablers for new aerospace architectures that create step changes in performance. In the early days of aircraft, wood, fabric, and iron were used, and the airframe and propulsion systems were subsequently optimized to maximize the performance of these materials. Then high strength aluminums and steels helped to bring about a revolution in aircraft architectures and systems, followed by titanium and carbon fiber composites. Already, these advanced materials have been implemented and optimized in aerospace structures over many decades, and the world is poised for another revolution that will bring about the next step change in aircraft and spacecraft capabilities.

Metal additive manufacturing has the unique ability to manufacture the material at the same time as the component, providing the potential for unprecedented levels of customization and optimization. The goal of this work was to advance the state-of-the-art in metal additive manufacturing by exploring the flexural behavior of optimized Ti-6Al-4V sandwich structure produced using the EB-PBF process. The objectives of the research were:

- Utilize Ti-6Al-4V sandwich structure to target aerospace industry applications, and design specimen core structures to be weight optimized based on the intended load case using Design for Additive Manufacturing best practices.
- Compare five different lattice core topologies with three different grading methods applied, and explore anisotropy in the build direction for each sandwich configuration.
- Size sandwich structures for 1) quality material, 2) topology driven flexural behavior, and 3) complete powder removal.

- Determine flexural behavior empirically in 4-point bend, guided by the industry standard ASTM C393, and then demonstrate experimental repeatability by testing a minimum of three of each specimen configuration

3.1 METAL SANDWICH STRUCTURE IN AEROSPACE

In aerospace, weight is often the most important aspect of aerospace structures. The higher the empty weight of an aircraft or spacecraft is, the more thrust is required to produce lift. This requires more fuel to meet the mission range, and more fuel means more weight. Now to carry more fuel, the loads on the structure are higher, thus the size of the structure has to increase, and so does the weight. Following this negative design spiral can eventually lead to an aircraft or spacecraft that cannot perform the intended mission of transporting a payload a desired distance. Therefore, the goal is often simple, reduce the weight of the structure as much as possible without compromising safety, and everything else gets better, e.g. fuel consumption, range, payload, cost, etc.

Of course, reducing weight is rarely as easy as it sounds. A traditional form of structure known as “skin and stringer,” is where a thin sheet forms the skin with one side acting as an external aerodynamic flow surface, while a long stiffener is attached to the other side to help the skin maintain its shape under complex loading conditions. The intent of the stiffening stringer is to resist bending loads, and therefore it often takes the cross-sectional beam shape of a T, I, Z, or perhaps a closed hat stiffener. As was explained in Section 2.2.3, the second area moment of inertia is a measure of how well a beam will resist bending based on its cross-sectional shape. The more mass that can be placed further away from the neutral axis, will provide a larger moment and increased stiffness. This is why I-beams have tall thin webs connecting two perpendicular regions

of mass; however, increasing the distance adds weight so a trade-off happens between how much bending resistance is required versus how much the structure weighs.

Metal sandwich structures serve the same purpose of providing increased bending stiffness, and they have the added benefit of potentially having two smooth flow surfaces on either side. They also take up less internal volume without the need for long back-side stiffeners. A skin and stringer architecture can run into problems when placed around hot environments created by a turbine or rocket engine. For example, in the case of a fairing around the turbine section of a commercial turbo-fan engine, the external flow surface requires a smooth skin. Beneath the skin is an arrangement of stiffeners that extend inwards towards the hot turbine case. The closer the ends of the stiffeners get to the case, the higher the temperature of the structure, which can degrade the material capability and reduces the intended stiffness. Growing the length of the stiffener further exacerbates the problem of increasing the temperature of the structure and the weight, without much benefit. A solution then is to move the flow surface further away from the turbine case to get the required stiffness without the temperature degradation, but that results in added wetted area (i.e. drag) and an increased radius which means more material (i.e. weight).

This is merely one situation in which metal sandwich structures can play a critical role. Similar structural stiffnesses can be obtained, with less depth of volume. The external flow surface can be placed, closer to the turbine, reducing the wetted area of the airframe, while maintaining sufficient clearance to the hot engine case. Therefore, a hollow metal sandwich panel sounds like a great solution, at least until it comes time to manufacture one. Questions that come to mind may include: How do you form the face sheets? How do you form the core? How do you join them together? How do you resolve defects such as disbonds, voids, and core distortion? How do you attach it to surrounding components? Conventional forming and joining processes are often

challenged to produce even simple metal sandwich structures that meet contour, quality, strength, and cost requirements.

Metal PBF appears to have the ability to solve a lot of the current problems with sandwich fabrication. While not without its own limitations, suddenly a complex metal sandwich structure can be printed to shape without any forming or joining operations. Panel thickness can be highly tailored and continuously variable without adverse core distortion or core crush. The density of the panel can also be continuously varied to achieve local strength needs without having to splice together different core densities, and gauge thicknesses can change throughout. Interfaces and attach points can be built-in and customized to evenly distribute loads, reducing stress concentrations and structural sizing. No longer would a localized worst case loading condition have to drive the same strength requirement to everywhere else, e.g. increase the gauge thickness of a metal sheet, just to make the component producible. Thus, the motivation for studying new methods to produce metal sandwich structures exists, and they are applicable to aerospace and other industries.

It was shown that the vast majority of published literature has almost exclusively sought to obtain bulk properties of lattice structures through compressive tests. While compression strength is important for some applications, it is not as critical for sandwich panels which experience a combination of tension and compression loading in bending. The 4-point bend experiment was specifically designed to analyze the flexural properties of a material system. Thus, it can provide a more accurate representation of the behavior under bending loads, because the structure experiences a uniform bending moment, and a more constant stress distribution between the load points. Another benefit to using the 4-point bend test is that it is more sensitive to flaws and defects that might not be apparent in a compression scenario. Therefore, the flexural behavior of a

sandwich structure subjected to 4-point bend can provide more relevant data for design and analysis since it better simulates real-world conditions.

Titanium is common in aerospace structures due to its high specific strength, low density, high temperature resistance, hardness, and corrosion resistance. However, it is a more expensive material than aluminum, and all those qualities for which it is valued also make it difficult to work and shape into components, adding to the cost. Additive manufacturing of titanium alloys helps to circumvent many of those challenges while offering even more capability to increase performance of a structure. Since Ti-6Al-4V is one of those alloys often employed in aerospace sandwich structures, it was selected for this work, ensuring the results would be relevant to advancing the state-of-the-art.

3.2 MANUFACTURING QUALITY

There has been little focus on the production of quality lattice structures for experimental evaluation, as noted throughout Chapter 2. While it is necessary to explore the manufacturing limitations of PBF processes, particularly feature resolution, there is little value in assessing the performance of a structure with defect driven deformation behavior. Such structures would not be useful in most industrial applications because the performance is low and unpredictable. Therefore, to advance the best practices in PBF process technology, a quality evaluation of manufactured specimens is required.

One of the cornerstones of any scientific investigation is repeatability of the results, because it demonstrates that a robust methodology was applied and adequate control was maintained over variables. In this way, it allows the results to be verified by other scientists and ensures they are consistent, not random or by error. The objective of testing a minimum of three of each specimen configuration helps to identify any errors or biases in the design of the experiment, and it enhances

the overall credibility of this research. Essentially, it adds to the accumulation of reliable knowledge from which future scientists can continue to build upon. When combined with future research into flexural properties of metal additive sandwich structures, repeatability will lead to generalizations that can be made about their behavior for deployment in industry.

For EB-PBF, applying the objective of complete powder removal was critical. Because the flexural test specimens undergo combined compression and tension loading, sintered powder embedded within the structure could provide artificially stiffened load paths and change the overall behavior. There would also be significant variability between the specimens of the same configuration as it would be impossible to guarantee that the same amount of powder remained in all the same places. Previous studies had demonstrated the difficulty of removing unmelted powder depending on the unit cell size and the relative density. Therefore, the specimen design had to be such that powder removal could be performed by the PRS, and verified by visual inspection and mass measurement.

Anisotropic behavior is nothing new in many manufacturing processes. The most notable being the layering and directionality of fiber reinforced polymer matrix composite materials, though it is much less discussed when it comes to metals. Traditionally, metals are often assumed to be isotropic, even though fabrication processes such as rolling can create rolling direction dependent microstructures and residual stresses that result in some anisotropy. In additive manufacturing, build direction dependent anisotropy is well documented and therefore must be considered for this EB-PBF investigation. To meet this objective, the experiment included specimens built in the horizontal (primary) and vertical (secondary) directions. This also demonstrated the importance of build orientation selection when designing components for EB-PBF.

Quality material characteristics are important to the repeatability of experimental results, and to the comparison of topological optimization methods. Prior to the flexural specimen builds, a series of trial builds were conducted to determine the optimum build orientation, lattice topologies, unit cell size, and relative densities. These trials served to narrow the scope of the investigation to designs that would result in high quality, lightweight structures, suitable for EB-PBF, and representative of industrial applications. Careful consideration was required for how the face sheets and lattices would best be resolved and integrated, as well as to ensure all sintered powder could be removed from the internal structure.

3.3 TOPOLOGY OPTIMIZATION

In the early days of structural topology optimization, it struggled for relevancy. All too often the solutions it produced were difficult to translate into CAD tools, and either expensive or impossible to manufacture by conventional processes. Some optimization methods for stiffener placement and gauge thicknesses made their way into early industrial applications, but most languished in the idealized theoretical realm. The birth of additive manufacturing technologies has created a resurgence in the field of topology optimization, because now the solutions can be easily produced and verified experimentally. For the design of lightweight metallic structures, employing optimization methods is essential to demonstrate the most advanced capabilities in the field.

Topology optimization is a powerful tool for AM, because it helps to maximize material efficiency, enhance performance, reduce the cost of design iteration, and creates innovative complex features that may have been previously unimagined. By combining AM with topology optimization, novel design solutions can lead to breakthroughs in the development of new products. To demonstrate novel sandwich structures that exemplify the potential for additive manufacturing, the objective was set to have lattice cores designed such that they could not be

produced any other way. Taking this direction helps align the results with emerging research trends and practical industry needs.

Due to the complexity that EB-PBF allows, the creation of customized parts tailored to specific requirements and applications is enabled. Therefore, the objective to optimize the specimens for the 4-point bend load case helps to showcase the significant advantages of bespoke solutions. Such DfAM solutions provide one of the key advantages to adopt PBF processes, despite lower technology readiness, and perhaps higher costs compared to conventional fabrication techniques.

The objective to compare different structural topologies is essential for evaluating performance, and to the understanding of material properties and mechanics. Various arrangements of structural components, particularly lattice cores within a sandwich panel, can exhibit different performance characteristics under the same loading and environmental conditions. A comparison can allow engineers to choose the topology that best fits their requirements, and helps to mitigate risk by identifying multiple potential failure modes. It also provides a breadth of exploration that can lead to the discovery of creative solutions or improvements to existing, and future, engineering problems. For this same reason it was necessary to include different grading methods that showcase the potential enhancement topology optimization methods can provide.

It is important to acknowledge the diverse learning dimensions embedded within the research objectives of this work. The methodologies applied herein have not previously been utilized for metal sandwich structures manufactured through EB-PBF, nor have they been systematically documented in the scientific literature. In the absence of established guidelines for the design, analysis, optimization, and fabrication of these novel Ti-6Al-4V sandwich specimens, many of the

approaches and techniques detailed in the following chapters were developed specifically for the purposes of this investigation. As such, the experimentally characterized flexural behavior of the optimized EB-PBF produced Ti-6Al-4V sandwich structures constitutes a meaningful contribution to the field of structural engineering and enriches the global scientific knowledge base.

Chapter 4. MATERIALS AND METHODS

This chapter outlines the materials, procedures, and analytical techniques employed throughout the course of this research. A diverse array of methods was utilized, some aligned with established industry standards and executed in accordance with documented protocols, while others were novel and developed specifically to address the unique challenges presented by this investigation. In pursuit of scientific rigor and reproducibility, each step of the experimental workflow, including material selection, fabrication processes, data acquisition, and analysis, has been thoroughly documented. The objective of this chapter is to provide a comprehensive and transparent account of the research methodology, thereby enabling future researchers to replicate the study and validate its findings.

The fabrication process used in this work was EB-PBF, and all samples or specimens were produced in an Arcam A2X located at the University of Washington Mechanical Engineering Department. In general, all samples and specimens were made from the titanium alloy Ti-6Al-4V, though there were differences in where the feedstock powder material was procured. This research began with feedstock material characterization, powder handling methods development, and EB-PBF process parameter adjustments. Those were followed by manufacturing quality assessments and experiments in tensile and compressive mechanical behaviors. Once confidence had been established in the quality of the produced material and fabrication processes, the sandwich beams were manufactured and tested in 4-point bend to determine their flexural behavior and failure modes. Subsequent analysis of the failed specimens was carried out by optical microscopy and scanning electron microscopy (SEM).

Aside from the physical processes conducted in laboratories, a great deal of analytical and numerical work was also accomplished. The methods described include hand calculations of beam

analysis, and finite element models for simulation. A new implicit CAD modeling platform called nTopology was utilized for both the design and finite element analysis of samples and specimens. In order to have confidence in this new tool and the modeling methods, a cantilever beam problem was solved and the results checked with classical beam theory. Also, to ensure the proper boundary conditions for the 4-point bend scenario were correct, a simple solid beam was simulated, and the calculations of stress and displacement were validated. Those modeling methods and techniques were then applied to the successful design and analysis of the lattice sandwich beams.

4.1 ELECTRON BEAM POWDER BED FUSION

All specimens utilized in this study were fabricated using an EB-PBF system, specifically the Arcam A2X platform. This machine operates with a 60 keV electron beam as its energy source, which selectively melts designated regions of the powder bed in accordance with the digital build file. The melting process corresponds to successive cross-sectional layers of the intended geometry, deposited in a layer-by-layer manner with a nominal layer thickness of 0.050 mm. The following section details the procedures employed for build preparation, process monitoring, and post-build extraction of the manufactured components.

The procedure for setting up the machine to run a build is well documented in the Arcam A2X user manual, and therefore unnecessary to duplicate here. In general, the procedure begins with an inspection of the build chamber for cleanliness, and to identify any parts that may need to be adjusted or replaced. Then the powder hoppers were filled approximately equally, with the desired amount of powder. They were inserted into the machine using a Svenema trolley, and powder was released from the hoppers to fill the bottom of the build tank. A thermocouple for monitoring the build temperature protrudes from the bottom of the build tank and is placed on top of the powder bed, then a clean 210 x 210 x 10 mm 316L stainless steel build plate is carefully placed on top.

Since the build plate is only loosely resting on a bed of powder, a procedure for manually leveling it was performed, and was deemed level when the rake barely cleans the entire plate uniformly. This was considered the zero position, and it was locked into the machine programming.

Next, was the verification that the rake fetches the correct amount of powder, and is known as the “pencil test.” The machine operator adjusts the end stop position of the rake for each hopper such that a certain amount of powder falls over the top of the rake to form a pile about the size of a pencil (6-8 mm high) in front of the rake. It was also important to ensure the pile was uniform across the length from the front of the chamber to the back. While this procedure can be quite time consuming, requiring a great amount of patience, the machine does continuously adjust the fetch position throughout the build. To either side of the build area are holes where a small amount of powder is allowed to fall through, this is registered with proximity sensors below the build platform. The strength of the pulse tells the machine control software whether there is too much or too little powder, and subsequently makes small adjustments to the end stop fetch position. This feedback system also helps to indicate if a hopper is empty, thus the controller will pull powder from the opposite one so long as powder is still detected.

Following the establishment of fetch positions, the build plate was thoroughly cleaned of residual powder using a Nilfisk A15 vacuum system, and the requisite heat shield components were assembled and installed within the build chamber. Many of these internal chamber elements are classified as consumables due to progressive thermal degradation resulting from repeated exposure to elevated temperatures; thus, the preliminary inspection and replacement of worn components prior to each build cycle. To ensure the integrity of the vacuum seal, the chamber door and its mating surfaces were meticulously cleaned before closure.

Before the electron beam can be turned on for calibration, the build chamber was evacuated to a base pressure of approximately $1\text{e-}5$ mbar. Subsequently, a controlled backfill of helium gas to a partial pressure of $\sim 2\text{e-}3$ mbar was introduced to mitigate oxygen contamination and suppress electrostatic charging of powder particles. Maintaining a high level of cleanliness is critical to system performance, as achieving the target vacuum level can require several hours. Even minor leaks can prevent the chamber from reaching the desired pressure threshold, a condition that typically becomes apparent only after prolonged evacuation. In such cases, the chamber must be reopened, re-cleaned, and the evacuation process repeated, resulting in significant delays.

Alongside the physical setup of the build chamber was the setup of the machine control software, called EBM Control (version 5). This process began after the build file had been created through the use of Materialise Magics and the Arcam Build Assembler software. Magics is where STL models of the parts to be produced are arranged and oriented within the build volume, as well as adding any support features that may be required. The build model was then transferred to the Build Assembler where it was sliced into the build layers and translated into a readable format for the EBM Control. That build file was transferred to the A2X via USB and stored on the hard drive. The first step in setting up the EBM Control program involved restarting it, this reduced the risk of software errors occurring that could cause the build to fail.

The instructions for setting up the beam control software are also contained within the A2X user manual. The process began with selecting the build file, identifying the size of the build plate, and selecting the appropriate preheat theme. A “theme” is a collection of EB-PBF process parameter settings, and some were pre-defined (i.e. default) based on the machine manufacturer recommended values. In the case of the initial build plate Preheat theme, this defined settings such as the beam current (48 mA), speed (35,000 mm/s), scan area size (195 x 195 mm), focus offset

(80), and other parameters associated with pausing for heat shield outgassing, and time out criteria. Then each component within the build was assigned a theme that controls build speed and quality on a part-by-part basis. Despite the level of flexibility and customization this fabrication platform affords, for this work all samples and specimens were assigned the same theme, called the Melt theme, while any support features within the build volume were assigned the Wafer theme. The key difference between the Wafer and Melt themes is that the Melt theme provided a higher energy density and higher quality melt pool, whereas the Wafer theme uses a lower energy density to produce a weaker brittle material which can ease removal during post-processing.

The Melt theme process parameter set contained the settings specific to contour and hatching scans for each layer. The contour scan utilized a spot melting strategy with three passes, where the outer scan beam speed was 404 mm/s with a current of 5 mA, and the inner scans (assuming the scan area was of sufficient thickness) had a beam speed of 850 mm/s with a current of 10.2 mA. In regions that required an infill melt, a hatching scan used a continuous snaking scan pattern with a variable beam current (max 20 mA). The angle at which the hatching scan took place rotated dynamically, according to an internal function that was intended to limit the length of a scan line. During each layer, the machine software has four proprietary functions that continuously vary the hatching scan parameters in order to maintain melt pool quality. These are known as the Speed Function, Thickness Function, Turning Point Function, and Current Compensation Function. Each function had their own coefficients that determined their behavior in different situations.

All support features were assigned the Wafer theme parameter set, which employed a continuous line scan at a speed of 1600 mm/s with a current of 5.5 mA. When comparing the beam current settings between themes, it is important to note that the Wafer theme distributes the

scanning beam energy across the entire build area nearly simultaneously, while the Melt theme scan energy is much more concentrated into a smaller square region that sequentially jumps around to melt different sections of the layer. Therefore, the beam energy density of the Melt theme is much higher than the Wafer, even if the beam currents are similar. A full table of the various machine parameters for each theme utilized in this investigation can be found in Appendix C.

Upon achieving the target vacuum level, and following the assignment of component-specific build themes, the high-voltage power supply and electron beam were activated. Prior to initiating the build sequence, essential beam calibration procedures, that included beam centering and focus alignment, were conducted to ensure optimal performance. Beam centering was accomplished by visually aligning the electron beam to the geometric center of the build plate, which had been manually marked prior to chamber installation. Focus alignment involved observing the beam's flickering behavior, characterized by the appearance of two closely spaced alternating spots. Using keyboard-controlled adjustments, the beam parameters were tuned until the flickering ceased and the two spots merged into a single point. This alignment procedure was systematically repeated at nine discrete locations across the build plate to ensure uniform beam focus.

The accuracy of this calibration process is critical for build quality and repeatability. However, the procedure is inherently subjective, relying on the operator's visual acuity and interpretation of beam behavior, particularly at peripheral locations near the corners of the build plate, where visibility is limited. As a result, inter-operator variability may influence calibration consistency, and the sensitivity of build quality to these calibration nuances remains an open question.

Following the focus alignment procedure, the build was initiated. It began with the chamber pre-heat process using the parameters described in the build plate preheat theme. That process ran

until the thermocouple under the build plate measured 730 °C, taking approximately 30 minutes. At that point, the build program switched automatically to the layer-by-layer build process consisting of: 1) the raking of a new powder layer, 2) preheat 1 to lightly sinter the powder to a “jump safe” temperature, 3) preheat 2 to bring localized regions to a “melt safe” temperature, 4) full melting of the outer contours, 5) continued melting of the inner contours, 6) melting of the infill by a hatching scan, and finally 7) a post melt heating as needed to maintain a constant energy input per layer. Those operations continued according to the programmed themes until the final layer was completed.

For safety reasons, the build chamber door would remain locked until the build plate thermocouple reached 80 °C, which depending on the size of the build would take several hours. While maintaining the chamber vacuum levels, helium would continue to be circulated to help increase the rate of cooling. Once the chamber could be unlocked, pressure was restored to the chamber and the door was opened for build extraction. First, the powder hoppers would be removed and any excess powder would be recycled through the Arcam Powder Recovery System (PRS). Then, all loose powder inside the chamber would be vacuumed using the Nilfisk A15, which is ATEX classified for use with titanium powder.

The sintered titanium cake had to be manually lifted out of the build tank in order to transfer it to the PRS. This extraction process required locating the build plate within the cake by carefully chipping away sintered powder with a screw driver and the vacuum. Once all build plate edges were exposed, the build cake could be gently separated from the plate and placed in the PRS. Beneath the build plate, the thermocouple would be trapped within sintered powder and thus had to be cautiously excavated to avoid damaging it. Once all loose powder in the chamber had been

vacuumed, it was transported to the PRS to be filtered and mixed with the sintered powder. Any remaining powder from the hoppers was also unloaded into the PRS for mixing purposes.

Within the PRS, which is functionally analogous to an industrial abrasive blasting station, titanium powder was entrained in a compressed air stream to mechanically disaggregate sintered material and expose the underlying manufactured components. The operator actively manipulated both the orientation of the components, and the directionality of the air stream, to ensure thorough removal of residual sintered powder from all exposed surfaces. The entrained powder was subsequently routed through a filtration system designed to eliminate fine particulates, defined as those with diameters less than 45 μm .

Following complete powder removal from the components, the parts were extracted from the PRS, initiating the powder reclamation phase. The majority of the recovered powder accumulated in the base of a collection chamber and was easily transferred to a steel containment bin via removal of a drain plug. However, residual powder persisted within the internal transport conduits of the system. To facilitate near-complete powder extraction, the PRS was operated for an additional minute, directing the air stream toward all interior surfaces of the main chamber. This agitation mobilized remaining powder, which was then incrementally drained. The process was iteratively repeated until the air stream exhibited visual clarity, airborne particulates were no longer observed within the chamber, and manual tapping of the drain yielded no further powder discharge.

The powder collected from the PRS was processed through a Russell Sieve with a 150 μm mesh size. This step removed particles that were agglomerated, as well as any debris or contaminants from the feedstock. Once processed through the PRS and sieve, the titanium powder was either returned to the appropriate storage containers with desiccant, or placed back in the

machine hoppers for the next build. If more powder was needed to restore the quantity levels, new powder would be added to the PRS during the powder recovery of the previous build. The intent being that the new and used powder would be properly mixed and keep the powder distribution uniform. It is not recommended to simply place new powder on top of used powder in the hoppers, or vice versa, as that could lead to non-uniform part quality from the bottom of the build to the top.

The complete operational cycle of the Arcam A2X system, including machine setup, build extraction, and post-process restoration, required several hours per build. Given the complexity and number of procedural steps involved, adherence to standardized checklists was essential to ensure process fidelity and prevent omissions. Failure to execute any individual step could compromise build integrity, potentially resulting in part defects or complete build failure. Notably, regardless of build outcome, the time investment for setup, extraction, and restoration remained constant.

Maintaining a high standard of cleanliness was critical to process reliability, particularly when working with reactive materials such as titanium. A conservative approach was adopted, guided by the principle: “if it appears clean, clean it again.” To mitigate health and safety risks associated with titanium powder handling, appropriate personal protective equipment (PPE), including gloves, lab coats, safety glasses, and respiratory masks, was utilized consistently throughout all powder-related operations. It is important to note that the laboratory environment housing the Arcam A2X system lacked adequate control over key ambient conditions, such as positive pressure, humidity, and temperature. This limitation may have introduced variability in the operating environment over this works duration that spanned several years, and may have influenced certain aspects of process consistency and repeatability.

4.2 SUSTAINABILITY: REPURPOSE OF WASTE L-PBF POWDER

A key innovation in this study was the demonstration of the sustainability potential inherent in the EB-PBF process. Rather than utilizing virgin Ti-6Al-4V powder sourced directly from the equipment manufacturer, this research prioritized the repurposing of waste powder originally designated for L-PBF applications. This powder was obtained from a local source with several years of operational experience in L-PBF, and it had accumulated as “oversized” material through routine powder handling and sieving protocols. Due to its assumed PSD exceeding the acceptable threshold for L-PBF reuse, the powder was considered unsuitable for further use in laser-based systems and was slated for disposal.

Given the hazardous nature of titanium powder and the associated costs of compliant disposal, its redirection into EB-PBF applications represents a significant advancement in sustainable additive manufacturing practices. As discussed in Chapter 2, L-PBF systems typically require finer powder particles to ensure optimal layer deposition and energy absorption. However, in an EB-PBF system, these smaller particles are prone to undesirable phenomena such as smoking, characterized by electrostatic overcharging and erratic powder spreading, when exposed to the high-energy electron beam. As such, EB-PBF systems have been designed to be more tolerant of coarser powder fractions. This strategic reuse of otherwise discarded powder not only reduces material waste and environmental burden, but also underscores the adaptability of EB-PBF technology in supporting circular economy principles within advanced manufacturing.

Several additional steps were required in order to utilize the waste L-PBF powder in the Arcam A2X. First, the feedstock had to be analyzed to determine its “as-received” status, because the pedigree of the powder was unknown. There was no record of how many cycles any of the quantities had been processed through the L-PBF system, and since it had been set aside for

disposal, the quality of storage was uncertain. Once analyzed, powder handling procedures and tracking were adjusted to ensure this powder source did not mix with any others. Process parameters also had to be adjusted to avoid smoking the powder in order to build samples for mechanical characterization. It was eventually determined that the objectives of this research could be accomplished with the use of the waste L-PBF titanium powder. Thus, throughout this work, all builds were accomplished utilizing the waste powder as feedstock, unless otherwise stated.

4.2.1 *L-PBF Powder As-Received Analysis*

Due to the fact that the L-PBF machine cycling and storage conditions of the as-received Ti-6Al-4V powder were unknown, a chemical analysis was required to determine if there were any changes to the chemistry. A small quantity of powder was obtained from several containers and mixed into a single sample. That sample was analyzed three times using a Rigaku NEX CG Energy Dispersive X-Ray Fluorescence Spectrometer, with the powder being mixed between runs to redistribute the components. For x-ray spectrometric analysis, standardization was performed by also using Standard Reference Material (SRM) 654a, a certified Ti-6Al-4V standard material in the form of a disk, 31 mm in diameter and 6.4 mm thick. The powder values obtained from the spectrometer were compared with the values from the certified disk to validate the as-received powder was in fact Ti-6Al-4V. In addition, three samples were processed with a LECO inductively coupled plasma optical emission spectroscopy (ICP-OES) analyzer to determine if the quantity of Oxygen, Nitrogen, or Hydrogen were outside the expected values.

High flowability of powder is necessary in PBF processes to assist with the uniform distribution across the powder bed. Irregularly shaped particles do not flow well, and may clump or bridge voids in the powder bed creating defects. Therefore, the flowability of the as-received

powder was measured per ATSM B213 Standard Test Method for Flow Rate of Metal Powders, which describes the test method using a Hall Flowmeter funnel as displayed in Figure 4.1. The static flow method procedure called “Method 1” was followed where 50.0 g of powder was poured into the funnel with the lower orifice blocked. Timing the flow with a stopwatch began when the block was removed and then stopped the instant the last of the powder exited the orifice. After applying a correction factor, an average time for each sample was obtained following 5 runs, and three samples were used each time this procedure was performed. The average of the three samples was reported as the overall corrected Hall flow rate $FR_H = t \text{ s}/50 \text{ g}$, where t is the flow time in seconds to the nearest second. Calibration of the apparatus was accomplished per the standard procedure using Chinese emery; thus, it provided the correction factor that was applied to the flow times.

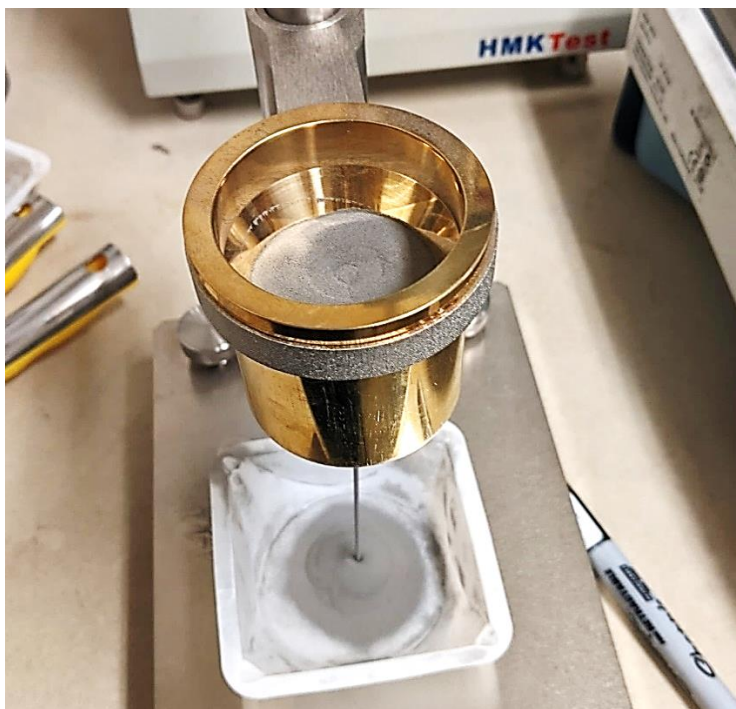


Figure 4.1. General method measuring flowability of powder using a hall flow meter per ASTM B213.

To characterize particle morphology and quantify the particle size distribution (PSD), including the D10, D50, and D90 diameter thresholds, scanning electron microscopy (SEM) was utilized in conjunction with image-based analysis techniques. Powder samples were imaged using a FEI XL30 Sirion SEM, with individual specimens affixed to the sample stage using conductive carbon tape to ensure positional stability and minimize charging effects during imaging. SEM micrographs were acquired at multiple magnifications to capture a representative range of particle sizes and surface features.

Subsequent image analysis was performed using a custom MATLAB R2023a script, detailed in the Appendix. The script leverages the `imfindcircles` function to detect and measure particle diameters based on circularity criteria. Detection sensitivity and size thresholds were controlled via three user-defined parameters: lowerbounds, upperbounds, and sensitivity. These variables were calibrated to optimize particle recognition across varying image resolutions, enabling accurate extraction of PSD metrics for comparative analysis.

Lowerbounds served as the smallest radius of pixels that the program detects as a circle. Any value less than or equal to 6 gives a warning of inaccuracy; thus, that value was set for all images analyzed. Upperbounds was similar to lowerbounds, except it was the maximum radius that a circle could be drawn. Sensitivity was a more complex function, but essentially the higher the number, the higher the threshold for what the program called a “circle.” All of those parameters were tuned until a best set was found that had the least amount of user interpreted error. Because each image contained an information bar at the bottom, an important value to adjust was the edge buffer scale (ebf), which set the distance from the bottom of the image that `imfindcircles` searches for circles.

The program also included the function `ginput` in order to convert pixel lengths into physical measurements. When running the analysis, the operator was prompted to click either side of the scale, and then enter the scale value into the prompt. The function retrieves the x distance in pixels between the click locations to determine image lengths, and then compiles all of the circle sizes into arrays. Since it was important to display the full range of particle sizes detected, no outliers were removed from the reported data set. The MATLAB source code for this analysis can be found in Appendix A.

This image analysis method while efficient, is not without sources of error. The most obvious being the user having to click the ends of the scale bar, where it is very difficult to pick the exact same locations every time and accuracy would vary from person to person. Inaccuracies were found within the `imfindcircles` function where it could not always identify every single powder particle visible. In some cases, the lack of contrast was a contributing factor, particularly particles in the background, because `imfindcircles` considers the contrast of a shape versus its background to determine whether it is a circle. Particles in the background could be partially obscured by shadows, making them less likely to be picked up by the algorithm. Also, `imfindcircles` can only find circular objects; thus, irregularly shaped oblong particles would not automatically be detected and simply adjusting the sensitivity had no effect. Oversensitivity, when the sensitivity value of `imfindcircles` was too high, caused errors where the software detects larger circles that encompassed preexisting ones.

4.2.2 *Powder Reuse Tracking*

Powder handling and tracking procedures are key to quality assurance and repeatability for any powder additive manufacturing operation. Throughout the EB-PBF build experiments, the same waste L-PBF feedstock powder supply was utilized, and it was carefully segregated from others to

reduce the risk of contamination. After an amount of powder was extracted from the supply, it was weighed to the nearest pound by a Toledo No Springs Honest Weight Dial Floor Scale with a 1000 lbs capacity. The powder was held within the plastic shipping containers, which also contained various amounts of desiccant, and were placed on the scale shown in Figure 4.2. The extra weight of the containers and the desiccant did not account for more than a pound and was considered negligible based on the measurement equipment and required accuracy. Once the scale reading stabilized, an image of the reading was taken to record the result.



Figure 4.2. General method for weighing powder quantities.

Due to the unknown pedigree of the Ti-6Al-4V powder, any time a quantity of powder was extracted from the supply, it was cycled through the PRS 3 to 4 times prior to any EB-PBF experiments. A single cycle consisted of pouring a small amount of powder into the PRS chamber, taking care to avoid completely blocking the intake collector at the bottom, then spraying the nozzle around the chamber for about 30 seconds until most of the powder in the bottom was gone. Then more powder was added, in small increments, until the entire new batch of containers were empty. The powder was then removed from the PRS, sieved, returned to the containers, and weighed. This procedure was necessary to reduce the number of fines contained within the L-PBF feedstock, as well as remove any oversized contaminates. That powder was then deemed suitable for use in the Arcam A2X.

During the course of running experiments the quantity of powder available would slowly be reduced. Sources of consumption were: it being melted into specimens, failed builds, spillage from handling, or filtered out by the PRS or sieve. When the total from the used batch became too low to sufficiently fill the hoppers, more powder would be pulled from the supply and processed through the PRS a minimum of three times. Then the remaining amount from the used batch would slowly be added to the PRS with the newly processed powder in order to mix two into a new batch. Each time powder was removed from the PRS and sieved, it was placed in containers and weighed to record how much powder was lost each time it was cycled. Throughout the process of conducting this research, three batches were created (in sequence) totaling 314 lbs (142.43 kg) of titanium powder consumed from the waste L-PBF supply. A table of the batches and weights that were recorded can be found in Appendix A.

4.2.3 *Process Parameter Adjustment*

The default process parameters provided by the EB-PBF machine manufacturer are designed for productivity with the powder feedstock they supply. This means those parameters may not be conducive with powder from alternative sources, and would require some adjustment. With a smaller PSD powder intended for a L-PBF machine, it still had higher sensitivity to smoking even after multiple passes through the PRS and sieve. An initial trial build was conducted to test the waste L-PBF powder under default EB-PBF process parameters, and changes were based upon those results.

The initial trial build was intentionally brief and based on a geometry that had previously been fabricated successfully using standard EB-PBF feedstock. However, despite multiple prior successes, two consecutive build attempts using the repurposed L-PBF powder failed at the same stage, immediately following the onset of beam-induced preheating of the build plate. For these trials, approximately 50 kg of the processed waste L-PBF Ti-6Al-4V powder was loaded into the hoppers. The left and right hoppers were each filled to approximately 50% capacity, and the machine was configured in accordance with the manufacturer's operational guidelines. Notably, the vacuum evacuation process required an atypically prolonged duration of approximately four hours. During this time, powder migration was observed along the periphery of the build plate, as illustrated in Figure 4.3. Despite this anomaly, the chamber ultimately reached the target vacuum conditions, allowing the build sequence to proceed.

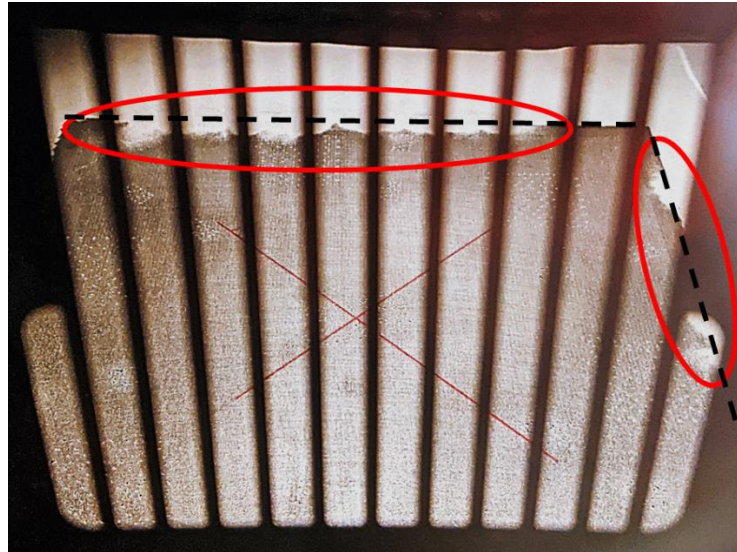


Figure 4.3. Titanium powder migrated back onto the build plate during the machine setup vacuum procedure.

Immediately upon initiation of the build plate preheating sequence, the system registered a series of arc trip events and recoverable faults. In each instance, the electron beam was activated, an arc trip was detected, and the beam was subsequently shut down. Approximately 15–20 such interruptions occurred within a span of several minutes, ultimately leading to an automatic termination of the build process. Visual monitoring through the observation window was severely impeded due to the presence of a dense particulate cloud, intermittently illuminated by incandescent titanium powder particles propelled upward. Upon opening the chamber, extensive powder dispersion was observed throughout the interior, including deposition atop the heat shield and in locations where powder accumulation had not previously been recorded post-build.

Furthermore, the region surrounding the build plate exhibited significant powder depletion, particularly along the periphery, suggestive of smoking phenomena as depicted in Figure 4.4. The build plate itself was uniformly coated with a fine layer of powder, obscuring the central alignment mark (X). A thin layer of titanium particulates was found adhered to the column protection foil

and the grounding ring securing it; both components were subsequently replaced. At the time, it remained unclear whether the observed anomalies were attributable to the characteristics of the recycled L-PBF powder or to a malfunction within the machine. Consequently, the chamber was thoroughly cleaned, and the system was reset for a second build attempt. However, the same failure mode was replicated.

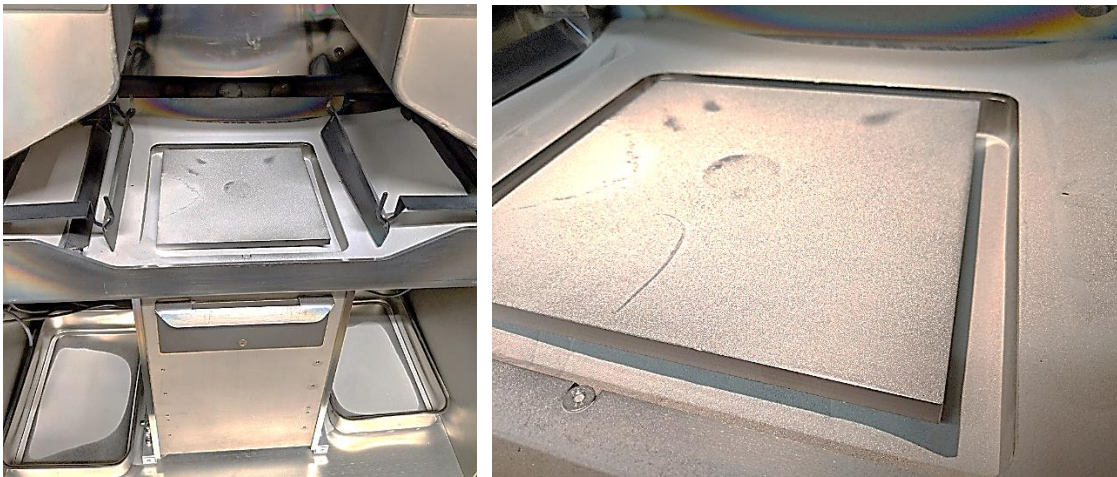


Figure 4.4. Arcam A2X build chamber (left) following the initial build with the waste L-PBF powder, and a close up of the powder covered build plate (right) with no powder around the edges following multiple smoke events.

As outlined in Section 2.1.2, the predominant mechanism underlying powder “smoke” events is attributed to electrostatic repulsion among powder particles. It was observed that such events may be initiated even in the absence of direct electron beam interaction, specifically during the preheating phase prior to powder deposition onto the build plate. One hypothesis posited that the thermocouple situated beneath the stainless-steel build plate may have failed to provide adequate electrical grounding, thereby allowing electrostatic charge accumulation. However, subsequent inspection confirmed that the thermocouple maintained effective contact with the build plate and was operating correctly. An alternative explanation considered the possibility of the residual

powder contamination along the build plate edges, which could potentially have triggered a smoke event. To mitigate this variable in later builds, a shallow trench was imposed around the perimeter of the build plate to isolate the build area. This intervention successfully eliminated edge powder interference, indicating that it was not the primary cause of the observed phenomenon.

Investigation into the machine parameters, which were optimized for the manufacturer's powder and productivity, it was noted that the preheat beam current was set to the maximum 48 mA in order to heat the chamber as quickly as possible. Therefore, another trial was initiated with the purpose of slowing down the preheating ramp rate. A new preheat theme was created where the only change was reducing the beam current to 1 mA. When the start plate preheat began at 1 mA, the powder did not smoke, but the build plate temperature rise was very slowly. After several minutes, the temperature had risen 5 °C, at which point the process was stopped and the plate preheat theme was revised to 10 mA and restarted. The temperature rose to 80 °C, and the process was stopped again to increase the current to 20 mA. With no smoke events, and the build plate temperature increasing to 133 °C, the cycle was continued each time upping the current by 5 mA and checking for smoke.

Ultimately, when the beam current was set to 45 mA and the build plate temperature reached 648 °C without triggering a smoke event, the process was intentionally halted to implement the default preheat theme at 48 mA. Under the standard configuration, the machine successfully completed the build plate preheating phase, achieving the target temperature of 730 °C. The build proceeded to completion without further modifications to the layer preheat or melting parameters, and no powder smoke events were observed. Due to the presence of the shallow trench around the perimeter of the build plate, the initial layers exhibited reduced powder coverage. However, after

several layers, the powder raking process normalized, and no adverse effects associated with this procedural adjustment were identified.

Notably, the preheating phase concluded at a beam current level that had previously induced powder smoke at ambient temperature. This observation aligns with findings in the literature indicating that elevated temperatures enhance particle stability and mitigate susceptibility to electrostatic overcharging. Consequently, the critical question for this recycled L-PBF powder source became: what is the maximum allowable preheat beam current that can be applied without initiating a smoke event? This revised temperature ramping strategy was adopted for all subsequent builds utilizing the waste L-PBF powder. Although, due to the manual nature of the current adjustments, slight variations in timing and temperature thresholds were inherent across builds.

4.2.4 *Mechanical Characterization*

A mechanical analysis of material produced with non-standard powder feedstock is an important component of material characterization. The mechanical properties provide insight as to the quality of the waste L-PBF Ti-6Al-4V and whether it was comparable to those published in industry and literature. Tensile and compression experiments were performed on a Instron 5585H (250kN capacity) machine with Instron mechanical wedge grips (100kN capacity), an Instron 2525-174 load cell (250 kN capacity), and an Instron 2630-101 (10 mm gauge length) extensometer that spanned the middle of the tapered section of tensile coupons.

Round ASTM-E466, 9 mm diameter tensile specimens were manufactured from the waste L-PBF Ti-6Al-4V powder. This specimen design, with its thick tapered circular cross section, was selected as being more representative of AM geometries. They were all printed using the Melt Theme process parameters in the vertical orientation, parallel to the build z-direction, which is

well known to be the direction of least strength. With guidance from ASTM E8 “Standard Test Method for Tension Testing of Metallic Materials” the coupons were tested under uniaxial monotonic loading, without any post-processing (i.e. as-built condition), as seen in Figure 4.5.



Figure 4.5. ASTM-E466, 9 mm dia. tensile specimen in Instron 5585H loading grip with Instron 2630-101 extensometer attached.

Two sets of tensile specimens were produced. The first set was an initial sampling of nine, built in the central region of the build envelope, and those preceded the production of any 4-point bend sandwich lattice specimens. Average tensile properties from that first set, such as yield strength and modulus of elasticity, were used for finite element analysis and structural sizing optimizations. The second set was conducted following production of all 4-point bend sandwich lattice specimens. It contained 35 tensile specimens, evenly distributed throughout the build area, and they corresponded with locations of sandwich lattice beams. This was intended to help determine if there were build area location effects that might explain any results seen from the sandwich specimens, as well as ascertain if the material characteristics had changed over the course of the experiments.

Compression experiments were conducted using the same Instron machine with smooth platens instead of grips, and loading was conducted either to specimen failure or maximum machine capacity (250 kN), whichever came first. An initial set of non-standard compression samples were 12 mm tall cylinders that had been extracted from the grip sections of the ASTM-E466 specimens. A second set of compression coupons followed ASTM E9 “Standard Test Methods of Compression Testing of Metallic Materials at Room Temperature” guidance for medium cylindrical specimens of 13 mm in diameter and 38 mm in length. They were built alongside the second set of tensile coupons that followed the production of all sandwich beam specimens. All compression specimens were tested in the as-built condition with no post-processing.

4.3 MANUFACTURING QUALITY ASSESSMENT

Following the successful validation of the non-standard Ti-6Al-4V feedstock for use in the Arcam A2X EB-PBF system, a manufacturing quality assessment was undertaken to qualitatively evaluate the printability and structural integrity of various lattice configurations. The objective of this assessment was to identify geometries that could be reliably fabricated with acceptable quality using the repurposed powder, thereby establishing a foundational understanding of design constraints and process capabilities. Given limitations in time and resources, a comprehensive evaluation of all possible lattice architectures was not feasible. Instead, a targeted down-selection approach was employed to focus on a subset of representative geometries deemed most relevant to the research objectives. The selection and evaluation framework was organized into four primary criteria: 1) manufacturability within the EB-PBF process, 2) demonstration of additive manufacturing-specific design principles, 3) structural characteristics, and 4) relevance and precedence in existing literature.

The manufacturability criterion focused on evaluating whether a given lattice structure could be successfully fabricated using the EB-PBF process, and whether the resulting build quality would be sufficient to ensure that subsequent mechanical testing outcomes would not be unduly influenced by process-induced defects. Particular attention was paid to overhanging features, which were assessed for excessive surface roughness, the presence of visible defects, and the complexity of post-processing requirements, specifically the effort required for powder removal using the PRS. The demonstration of AM criterion emphasized the selection of lattice geometries that exemplify the unique capabilities and design freedoms enabled by AM technologies. Preference was given to structures exhibiting complex, organic morphologies that are infeasible to produce via conventional manufacturing methods, and which offer potential for multifunctionality, thereby showcasing the transformative potential of AM.

Under the structural performance criterion, emphasis was placed on achieving low relative density to promote lightweight design, while maintaining adequate mechanical strength. Additionally, the ability of the lattice to integrate seamlessly with face sheets was considered critical, as poor interface design could lead to stress concentrations and compromise structural integrity. Lastly, the literature relevance criterion prioritized lattice geometries that are well-documented in existing academic and industrial research. This facilitated comparative analysis and contextualization of results within the broader body of knowledge, enhancing the interpretability and impact of the findings.

To facilitate the design of face sheets integrated with parameterized lattice cores, the platform nTopology was employed. This software leverages implicit modeling techniques to define 3D geometries, enabling the representation of highly complex structures with significantly reduced computational power compared to traditional explicit systems. In addition to its advanced

modeling capabilities, nTopology offers native support for mesh generation, finite element analysis (FEA), and topology optimization, making it a comprehensive tool for the development and evaluation of architected materials. As with many advanced CAD and FEA platforms, nTopology presents a substantial learning curve, particularly when executing high-level operations such as topology optimization for sandwich panel configurations. To address this, a progressive "crawl-walk-run" implementation strategy was adopted. This methodology began with the generation and assessment of small-scale lattice samples during the manufacturing quality evaluation phase, thereby establishing a foundational understanding of the software's capabilities and limitations prior to engaging in the more complex design tasks.

The quality assessment was to examine both strut and TPMS type lattices core geometries. Within the TPMS category, both skeletal and sheet types were considered. Initially, a variety of these lattice types would be configured into 25.4 mm cubes with two opposing sides given face sheets that were designed to be 0.8128 mm (0.032 inches) thick. The number of unit cells per cube depended upon the lattice type, but ranged from 1 unit cell, to 8 unit cells in a 2^3 configuration, or 27 unit cells in a 3^3 arrangement. The relative density of each sample was calculated based on the CAD model of the lattice core and excluded the face sheets. Where the lattice core geometries interfaced with the face sheets, a blend radius of 0.8 mm was maintained. The thickness assigned to the lattice configurations depended on the geometry and ranged from 0.7 mm to 1.5 mm

This task was also to assist in determining the most suitable orientation within the print volume to manufacture the lattice cores with face sheets, as well as to establish an effective configuration of external support features. Initial trials involved orienting the face sheets parallel to the build plane, with support structures applied to the lower face sheet to facilitate heat conduction and mitigate warping. Subsequently, the build orientation was modified by rotating

the specimens 90 degrees, positioning the face sheets perpendicular to the build plane (parallel with the build direction) and attaching support features to the bottom edges of both the face sheets and the lattice core. Alternative orientations at intermediate angles, were considered if neither of the primary configurations yielded satisfactory manufacturing quality. Importantly, no internal support structures were introduced within the lattice cores to avoid post-processing complications. The manufacturing quality assessment was considered complete once viable sandwich structure configurations were successfully fabricated and deemed suitable for subsequent mechanical characterization via 4-point bend experiments.

4.4 SANDWICH 4-POINT BEND EXPERIMENT

In the existing body of literature, the mechanical characterization of lattice structures has predominantly relied on uniaxial compression testing to extract bulk material properties. While compression tests are relatively simple to implement and yield useful data for certain applications, they are insufficient for evaluating the performance of lattice-based sandwich structures intended for aerospace use, where bending loads are the primary mode of mechanical stress. Such structures are typically subjected to combined tensile and compressive forces along their span, necessitating a more representative testing methodology.

To better simulate operational loading conditions and obtain relevant flexural behavior, the 4-point bend test was selected. This method offers distinct advantages over 3-point bending, particularly in its ability to generate a constant bending moment and a more uniform stress distribution between the two loading points. This uniformity facilitates a more accurate assessment of flexural behavior, especially in architected materials with non-homogeneous and non-uniform cross sections. In contrast, 3-point bend tests concentrate stress at the central loading point, often resulting in localized failure that may be adequate for homogeneous solids, but it is

less informative for porous sandwich structures. For these complex geometries, the 4-point bend configuration allows failure to initiate at any location within the constant moment region, thereby increasing the likelihood of revealing latent defects or structural weaknesses that might otherwise remain undetected.

Following the mechanical characterization and manufacturing quality assessment, the Ti-6Al-4V sandwich beams were designed for flexural strength and failure mode characterization. The beam configuration followed guidance from ASTM C393 “Standard Test Method for Core Shear Properties of Sandwich Constructions by Beam Flexure.” This standard describes the short beam, 4-point (third span) loading configuration that was employed in this investigation. The beams had 0.8128 mm (0.032 inches) thick face sheets with 25.4 mm (1 inch) core thickness. They were 192 mm long, 48 mm wide, and the cores had a relative density of approximately 10% with final unit cell geometry determined by the manufacturing quality assessment.

Five lattice core configurations were selected from the manufacturing quality task. Using nTopology, the beam CAD models were designed to have three different core thickness gradients per lattice configuration, making a total of 15 unique sandwich beam configurations. A uniform thickness lattice served as a baseline for comparison, while two other functional grading methods were driven by either a density field or a stress field. The density field was generated from a SIMP based compliance minimization topology optimization of the lattice in simulated 4-point bend loading configuration. The stress field was based on the von Mises stress result from a FEA simulation of the 4-point bend experiment. Thicknesses of the lattice cores were bounded such that the minimum was set to 0.6 mm, which is near the minimum resolvable feature size, and the upper bound was varied to maintain the 10% relative density.

The final print orientation and support structure configuration were selected based on the outcomes of the manufacturing quality assessment. A single EB-PBF build was sufficient to fabricate the complete set of 15 distinct sandwich beam specimens. To enable a statistically meaningful evaluation of process variability, a minimum of three specimens per configuration were produced using the waste L-PBF Ti-6Al-4V powder feedstock. This approach ensured an adequate sample size for comparative analysis and mechanical characterization. For benchmarking purposes, an additional build was conducted using virgin Ti-6Al-4V powder supplied by the EB-PBF machine manufacturer. This reference build served as a control, allowing for direct comparison of structural quality and mechanical performance between specimens fabricated from recycled and virgin powder sources.

4.4.1 *Experimental Setup*

The experimental apparatus used to conduct the 4-point bend tests is shown in Figure 4.6. It consisted of an Instron 5585H (250kN capacity) load frame, with adjustable upper loading and lower support attachments. The lower support span was set at 142 mm, while the upper loading locations were set equidistant at 47.3 mm from each other, according to ASTM C393 description of 1/3 span loading configuration. Placed in line with the upper attachment was a 250 kN capacity Instron 2525-174 load cell, and to measure the flexural displacement, an Epsilon 3540 deflectometer was positioned at the center point of the lower face sheet. Roller pins of 10 mm diameter provided the smooth distribution of load across the width of the specimen and ensured good contact as the specimen deflected. The load frame displacement rate was set to 1 mm/min, and experiments were stopped at a flexure strain of 0.07 to avoid exceeding the capability of the specimen holding fixture.

While flexure strain was obtained directly through deflectometer measurements, the corresponding flexure stress σ_f was derived using classical beam theory in Equation (4.1):

$$\sigma_f = \frac{My}{I} \quad (4.1)$$

where flexure stress (σ_f) is a function of the bending moment (M), the distance from the neutral axis (y), and the second area moment of inertia (I) for a rectangular beam cross section. Alternatively, ASTM D6272 provides Equation (4.2) [197], which is specific to the third-span loading arrangement, as a simplified form of Equation (4.1):

$$S = \frac{PL}{bd^2} \quad (4.2)$$

where the stress at the outer edge (S) is a function of the support span (L), the measured load (P), the beam width (b), and the beam depth (d). However, both approaches are based on the assumption of a uniform solid cross section and overlook the varying density within the core structure as well as the fluctuating moment of inertia.

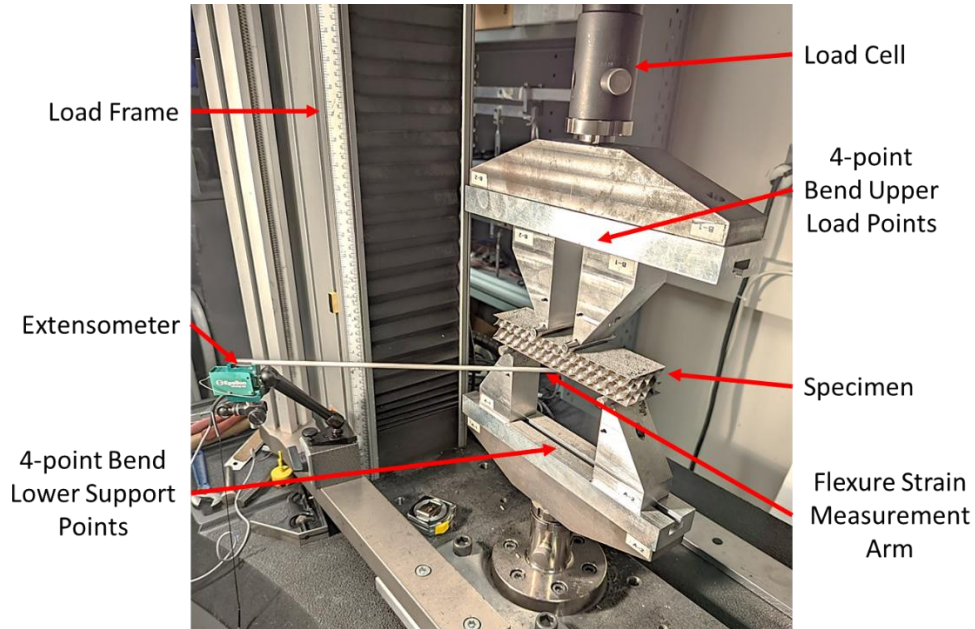


Figure 4.6. Experimental setup within the load frame of a lattice sandwich beam specimen in 4-point bend configuration with an extensometer to measure strain at the mid-span point.

To ensure experimental repeatability and measurement consistency, precise alignment of each specimen within the test fixture was critical. Upon extraction from the respective builds, specimens were systematically labeled to indicate their build origin (Builds 1, 2, 3, 4, V for vertical powder, and A for Arcam virgin powder), build orientation (denoted by an arrow), and core architecture (U: uniform, D: density graded, S: stress graded). Additionally, the “top” surface (T) was identified to establish the correct orientation of the face sheets relative to the load frame during mechanical testing. To standardize the loading conditions across all specimens, digital calipers were employed to accurately locate and mark the contact points for load application on the upper face sheet, support placement on the lower face sheet, and the central measurement location for extensometer attachment. Specimens were consistently positioned such that the build direction faced the front of the load frame; for vertically built specimens, this orientation was adjusted to face the left side of the frame to maintain uniformity in test setup.

It was desired during the flexural tests to capture any rapid crack propagation behavior; therefore, the setup included a high-speed imaging camera. The camera was a Vision Research Phantom v711 that has a range of 7,530-680,000 frames per second depending on the resolution. It was equipped with a f1.4 50 mm lens focused on the mid-span of the sandwich beam. Due to the high frame rate, additional light was needed to provide sufficient exposure, so a large lamp was positioned near the camera to illuminate the apparatus. The camera was set for continuous capture with looping memory that was able to store approximately 3 seconds of time. When a specimen experienced a sudden fracture, the camera was triggered manually to stop recording, which then saved the amount of time recorded prior to the trigger.



Figure 4.7. Flexure experiment setup showing with the high-speed camera in relation to the load frame.

4.4.2 *Hand Calculations*

In engineering analysis, it is essential to validate empirical observations and numerical simulations through classical analytical methods to ensure internal consistency and physical plausibility. Although analytical approaches may be limited in precision due to simplifying assumptions and reduced geometric or material fidelity, they serve as a valuable benchmark for interpreting results and reinforcing fundamental mechanical principles. This section presents a series of hand calculations based on classical beam theory and Hooke's Law, applied to a simplified representation of a sandwich beam subjected to four-point bending, as illustrated in Figure 4.8.

These calculations provide a foundational reference for assessing the mechanical response and verifying the trends observed in experimental data and numerical simulations.

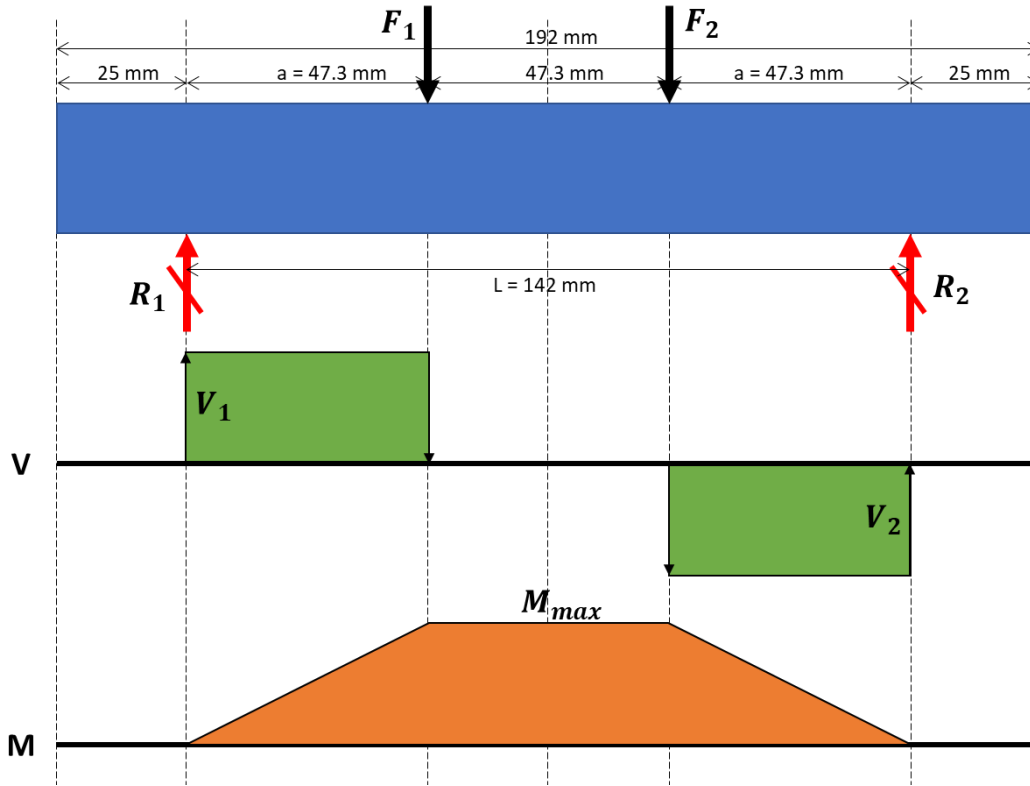


Figure 4.8. Illustration of the 4-point bend scenario showing locations of input and reaction forces, along with the accompanying shear and moment diagrams.

The deflection of a simple beam was calculated for various locations x along the length ($L = 0.142 \text{ m}$), which is actually the span distance between the reaction forces, not the full beam length. The force applied at each location $a = 0.0473 \text{ m}$ was assumed to be $F = 100 \text{ kN}$, and the material was assumed to be Ti-6Al-4V with a modulus of $E = 113.8 \text{ GPa}$. The simple beam had an assumed rectangular cross section where the width $w = 0.048 \text{ m}$ and the height $h = 0.0254 \text{ m}$. Therefore, the moment of inertia was calculated per Equation (4.3), and the deflections δ shown in Equations (4.4), (4.5), (4.6), and (4.7).

$$I = \frac{wh^3}{12} = 6.554 \times 10^{-8} \text{ m}^4 \quad (4.3)$$

$$\delta = \frac{Fx}{6EI} (3aL - 3a^2 - x^2) = 0.68 \text{ mm}, \quad (0 \leq x \leq a) \quad (4.4)$$

$$\delta = \frac{Fa}{6EI} (3Lx - 3x^2 - a^2) = 0.65 \text{ mm}, \quad (a \leq x \leq L - a) \quad (4.5)$$

$$\delta_{max} = \frac{Fa}{24EI} (3L^2 - 4a^2) = 1.36 \text{ mm}, \quad @ x = \frac{L}{2} \quad (4.6)$$

$$\delta_a = \frac{Fa^2}{6EI} (3L - 4a) = 1.18 \text{ mm}, \quad @ x = a \quad (4.7)$$

The slopes θ of the beam are calculated at various locations per Equations (4.8), (4.9), (4.10), and (4.11).

$$\theta = -\frac{F}{2EI} (aL - a^2 - x^2) = -0.026, \quad (0 \leq x \leq a) \quad (4.8)$$

$$\theta = -\frac{F}{2EI} (L - 2x) = 0.635, \quad (a \leq x \leq L - a) \quad (4.9)$$

$$\theta_1 = -\frac{Fa}{2EI} (L - a) = -0.030, \quad @ x = 0 \quad (4.10)$$

$$\theta_2 = +\frac{Fa}{2EI} (L - a) = 0.030, \quad @ x = L \quad (4.11)$$

The shear V and moment M calculations for this scenario are simple and displayed in Equations (4.12), (4.13), and (4.14).

$$V_1 = +F = 100000 \text{ N}, \quad (0 \leq x \leq a) \quad (4.12)$$

$$V_2 = -F = -100000 \text{ N}, \quad (L - a \leq x \leq L) \quad (4.13)$$

$$M = Fa = 4730, \quad (a \leq x \leq L - a) \quad (4.14)$$

The stress σ and strain ϵ at the lower mid-span location ($y = h/2$) was calculated according the Equations (4.15) and (4.16)

$$\sigma = \frac{My}{I_x} = \frac{4730 * 0.0127}{6.554 * 10^{-8}} = 9.16 * 10^8 \text{ Pa} \quad (4.15)$$

$$\epsilon = \frac{\sigma}{E} = \frac{9.16 * 10^8}{1.138 * 10^{11}} = 8.05 * 10^{-3} \quad (4.16)$$

In summary, the analytical calculations presented offer a foundational perspective on the mechanical behavior of beams in 4-point bend. While the simplified models do not capture the full complexity of the experimental configurations, such as localized deformation, material anisotropy, or geometric nonlinearity, they nonetheless provide valuable insight into expected stress, deflection, and strain behavior. The results serve as a useful baseline for comparison with numerical simulations, reinforcing the validity of observed trends in experimental data. This analysis contributes to a more comprehensive understanding of the structural response of a beam in 4-point bend, and it supports the broader findings of this investigation.

Building upon the insights derived from classical analytical methods, the subsequent section delves into more detailed evaluations of the sandwich beam specimens through numerical modeling. This approach enables the incorporation of complex geometries, material heterogeneities, and boundary conditions that are beyond the scope of traditional hand calculations. By comparing the analytical predictions with higher-fidelity simulations, a more robust understanding of the structural performance of additively manufactured sandwich structures was achieved.

4.5 NUMERICAL MODELING

Finite Element Analysis (FEA) is a computational technique used to simulate the mechanical response of structures by discretizing a continuous domain into a finite number of interconnected elements, collectively referred to as a mesh. When external loads are applied, and appropriate boundary conditions are defined, the resulting deformation and stress distribution are computed at

each node point, based on the assigned material properties that includes elastic modulus, Poisson's ratio, and density. In this research, FEA played a pivotal role in informing the optimization strategies for functionally graded sandwich lattice beams.

Due to the geometric complexity and spatial variation inherent in the lattice designs, the embedded FEA solver within the nTopology platform was utilized to perform simulations directly within the parametric design workflow. Although nTopology offers integrated analysis capabilities tailored for architected materials, it remains relatively new and less validated in industrial practice compared to established solvers such as NASTRAN, ANSYS, or ABAQUS. As a result, verification procedures were implemented to assess the reliability of the nTopology solver and to ensure the credibility of the simulation outcomes. These validation steps were essential for building confidence in the numerical modeling approach and for supporting the subsequent design decisions and experimental comparisons.

4.5.1 *Simple Cantilever Beam*

The initial verification involved a cantilever beam problem, selected to emulate the meshing strategy anticipated for the lattice core geometries. Specifically, the beam was discretized in a manner that reflected the meshing constraints of the thinner lattice regions, where only a single element may span the thickness. This condition is recognized in finite element literature as a potential source of numerical inaccuracies, particularly in capturing bending behavior and stress gradients.

While mesh refinement, achieved by reducing element size, is a conventional approach to mitigate such errors, the computational resources available for this study imposed practical limitations. Excessive refinement resulted in prohibitively large element counts and unacceptable solution times. Consequently, it was imperative to assess the solver performance under a

deliberately coarse mesh scenario, representing a “worst-case” condition with minimal through-thickness resolution. The objective of this test was to determine whether the nTopology embedded solver could yield sufficiently accurate results despite these constraints, thereby validating its applicability for the subsequent analyses of complex lattice structures within this research.

A simple cantilever beam was modeled in nTopology with a length of 50 mm, a width of 10 mm, and the thickness of 1 mm, with one end fixed. The beam was coarsely meshed with tetrahedron elements according to the workflow parameters shown in Figure 4.9, which resulted in an element count of 534. In order to turn an implicit body into a finite element mesh, it first requires rendering it with a voxel grid and then the voxel grid is turned into a surface mesh. However, the initial mesh conversion is very poor, and it required a remesh step that has more control parameters for refining the surface mesh and making it more uniform. Once there was an acceptable mesh that sufficiently captured the surface of the beam, it was converted into a volume mesh with a single element through the thickness.

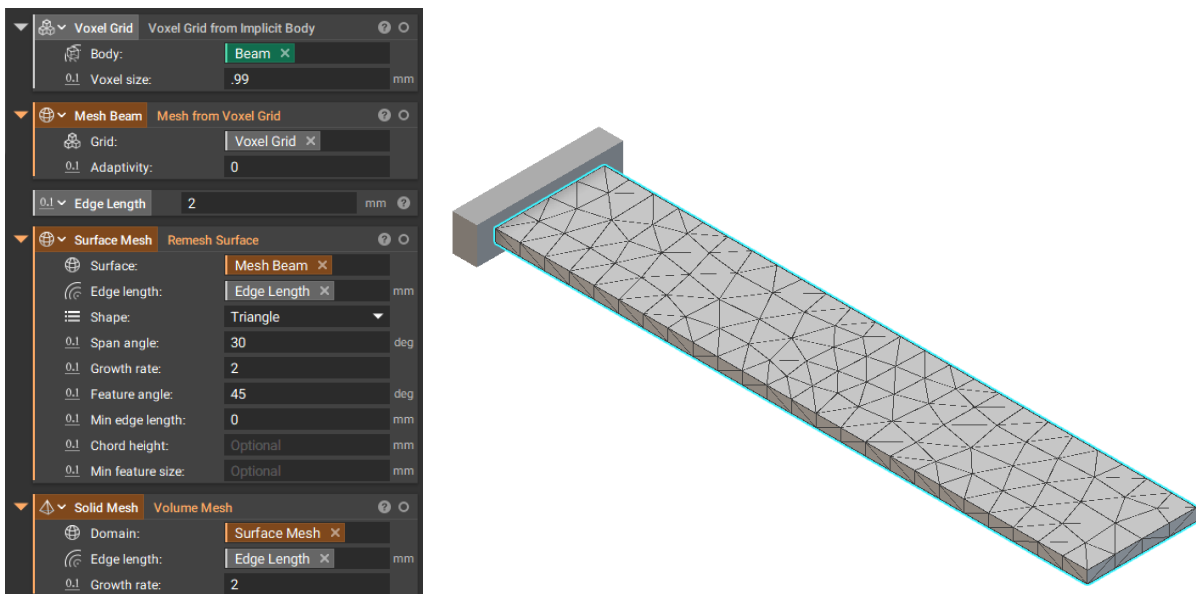


Figure 4.9. Model of a simple cantilever beam with a course mesh (right), and the model meshing parameters (left).

For the simulations, one end of the cantilever beam was fixed, such that all element nodes were not allowed displacement or rotation, while the opposing end was left free. Three different load application scenarios were setup to include an end load, a distributed load, and a torsion load. In bending, a 10 N force was applied across the width of the beam at the free end. The distributed load case had a 10 N force applied along the top surface across the entire width and length of the beam. In the case of torsion, a point moment of 1 Nm was applied along the width of the free end rotating about the length centerline axis. These scenarios are shown in Figure 4.10, with the following Ti-6Al-4V material properties applied: Young's Modulus of 113.8 GPa, Poisson's Ratio of 0.342, a density of 4.43 g/cm³, and a Shear Modulus of 44 GPa.

In order to verify the results of the FEM simulations, they were compared with hand calculations for displacement and maximum stress. Each simulation was performed twice, toggling between linear or quadratic elements, to determine which element type was more accurate and ascertain the relative error. Equations (4.17) and (4.18) were used to calculate displacement and stress for the simple bending load case,

$$\delta = \frac{FL^3}{3EI} = \frac{10 \cdot 0.050^3}{3 \cdot 1.1138 \cdot 10^{11} \cdot 8.333 \cdot 10^{-13}} = 4.39 \cdot 10^{-3} \text{ m} \quad (4.17)$$

$$\sigma = \frac{yM}{I} = \frac{(0.001/2) \cdot (10 \cdot 0.05)}{8.333 \cdot 10^{-13}} = 3.00 \cdot 10^8 \text{ Pa} \quad (4.18)$$

where the displacement (δ) is a function of the applied force (F) in Newtons, the beam length (L) in meters, the elastic modulus of titanium (E) in Pa, and the moment of inertia (I) in m^4 . The stress (σ) in Pa, is a function of the moment (M) in Nm, the distance the load is applied from the centerline (y) in m, and the moment of inertia (I). Similarly, the distributed load case hand calculations used Equations (4.19) and (4.20).

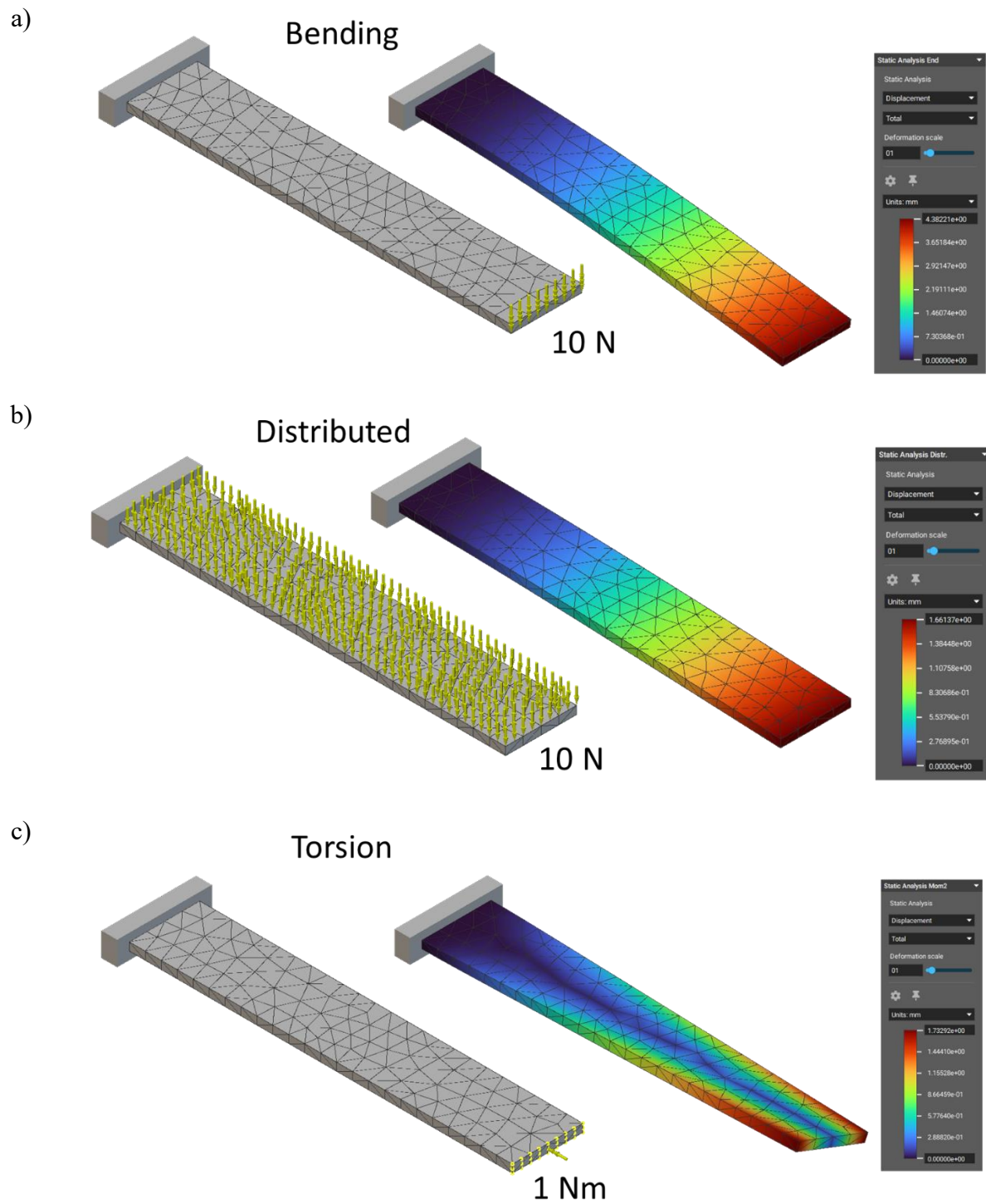


Figure 4.10. Finite element models with tetrahedron elements, showing displacement of a simple cantilevered titanium beam under three different loading scenarios: a) bending, b) distributed, and c) torsion.

$$\delta = \frac{FL^4}{8EI} = \frac{10 \cdot 0.050^4}{8 \cdot 1.1138 \cdot 10^{11} \cdot 8.333 \cdot 10^{-13}} = 1.65 \cdot 10^{-3} \text{ m} \quad (4.19)$$

$$\sigma = \frac{yM}{I} = \frac{\left(\frac{0.001}{2}\right) \cdot \frac{(10/0.050) \cdot 0.050^2}{2}}{8.333 \cdot 10^{-13}} = 1.50 \cdot 10^8 \text{ Pa} \quad (4.20)$$

In the torsional case, the angle of twist (θ , in radians), displacement (δ), and shear stress (τ_{max}) were calculated via Roark's [198] formulas, Equations (4.21), (4.22), (4.23), and (4.24), for a solid rectangular section, where T is the twisting moment (1 Nm), L is the beam length (0.050 m), G is the shear modulus of titanium (44 GPa), a is half the width (0.005 m), and b is half the thickness (0.0005 m).

$$K = ab^3 \left[\frac{16}{3} - 3.36 \frac{b}{a} \left(1 - \frac{b^4}{12a^4} \right) \right] = 3.12 \cdot 10^{-12} \text{ m}^4 \quad (4.21)$$

$$\theta = \frac{TL}{KG} = 0.364 \text{ rad} = 20.8^\circ \quad (4.22)$$

$$\delta = a\theta = 1.82 \cdot 10^{-3} \text{ m} \quad (4.23)$$

$$\tau_{max} = \frac{3T}{8ab^2} \left[1 + 0.6095 \frac{b}{a} + 0.8865 \left(\frac{b}{a} \right)^2 - 1.8023 \left(\frac{b}{a} \right)^3 + 0.9100 \left(\frac{b}{a} \right)^4 \right] \quad (4.24)$$

$$\tau_{max} = 3.20 \cdot 10^8 \text{ Pa}$$

Because the nTopology software provides results in von Mises stress, for a pure torsion application von Mises can be simplified following Equations (4.25), (4.26), (4.27). and (4.28).

$$\sigma_{vm} = \frac{1}{\sqrt{2}} ((\sigma_{11} - \sigma_{22})^2 + (\sigma_{22} - \sigma_{33})^2 + (\sigma_{33} - \sigma_{11})^2)^{1/2} \quad (4.25)$$

$$\sigma_{vm} = \frac{1}{\sqrt{2}} ((\sigma_{max} - \sigma_{22})^2 + (\sigma_{22} - \sigma_{min})^2 + (\sigma_{min} - \sigma_{max})^2)^{1/2} \quad (4.26)$$

$$\sigma_{22} = 0 \rightarrow \text{At top face sheet (plane stress)} \quad (4.27)$$

$$\sigma_{vm} = \sqrt{3}\tau_{12} \approx \sqrt{3}\tau_{max} = 5.55 \cdot 10^8 \text{ Pa} \quad (4.28)$$

Table 4.1 displays the results of the FEA models alongside the hand calculations with a percent error.

Table 4.1. Comparison of FEA stress results and hand calculations for the simple cantilevered beam in bending and torsion.

Load Case	Bending		Distributed		Torsion		
Measure	Displacement (mm)	Stress (MPa)	Displacement (mm)	Stress (MPa)	Twist (deg)	Displacement (mm)	Stress (MPa)
Analytical	4.39	300	1.65	150	20.8	1.82	555
FEA Linear error	0.65 85.1%	76 74.8%	0.26 84.3%	38 75.0%	1.9 91.0%	0.16 91.0%	132 76.2%
FEA Quad error	4.38 0.3%	302 0.8%	1.66 0.8%	144 3.7%	19.9 4.7%	1.73 4.7%	696 25.4%

The percent errors in Table 4.1 reflect what is generally well known, that linear Tet elements are overly stiff, and correlate poorly with analytical methods if an insufficient number of elements are used. Despite the mesh being relatively course, there was good agreement between the hand calculations and the quadratic tet results, with less than 5% error across the measured values in displacement and stress. The only outlier for the quadratic tet simulation was the torsion case von Mises stress, at 25.4%. Because there was good agreement from the angle of twist and displacement values, the stress result was investigated more closely.

Further inspection of the von Mises FEA plot in Figure 4.11a, revealed that the maximum stress reported was along the centerline at a node only one element away from the load application boundary condition. Stresses further toward the mid-span of the beam along the centerline, 3 to 5 elements from the load application boundary, were much lower at about 613 MPa, or 10.6% error. The course mesh showed that the centerline nodes were only two nodes away from the beam edge, likely causing the additional error. Upon refinement of the mesh in Figure 4.11b, the FEA von Mises stress in this region was reduced further to 560 MPa, resulting in only a 1.0% error from the analytical method. This was reinforced by the Mohr's circle, the S12 component, and the Min and

Max Principal Stress plots in Figure 4.12. These depict a pure shear situation in the same region that correlates well with the analytical shear stress of 320 MPa from Equation (4.24).

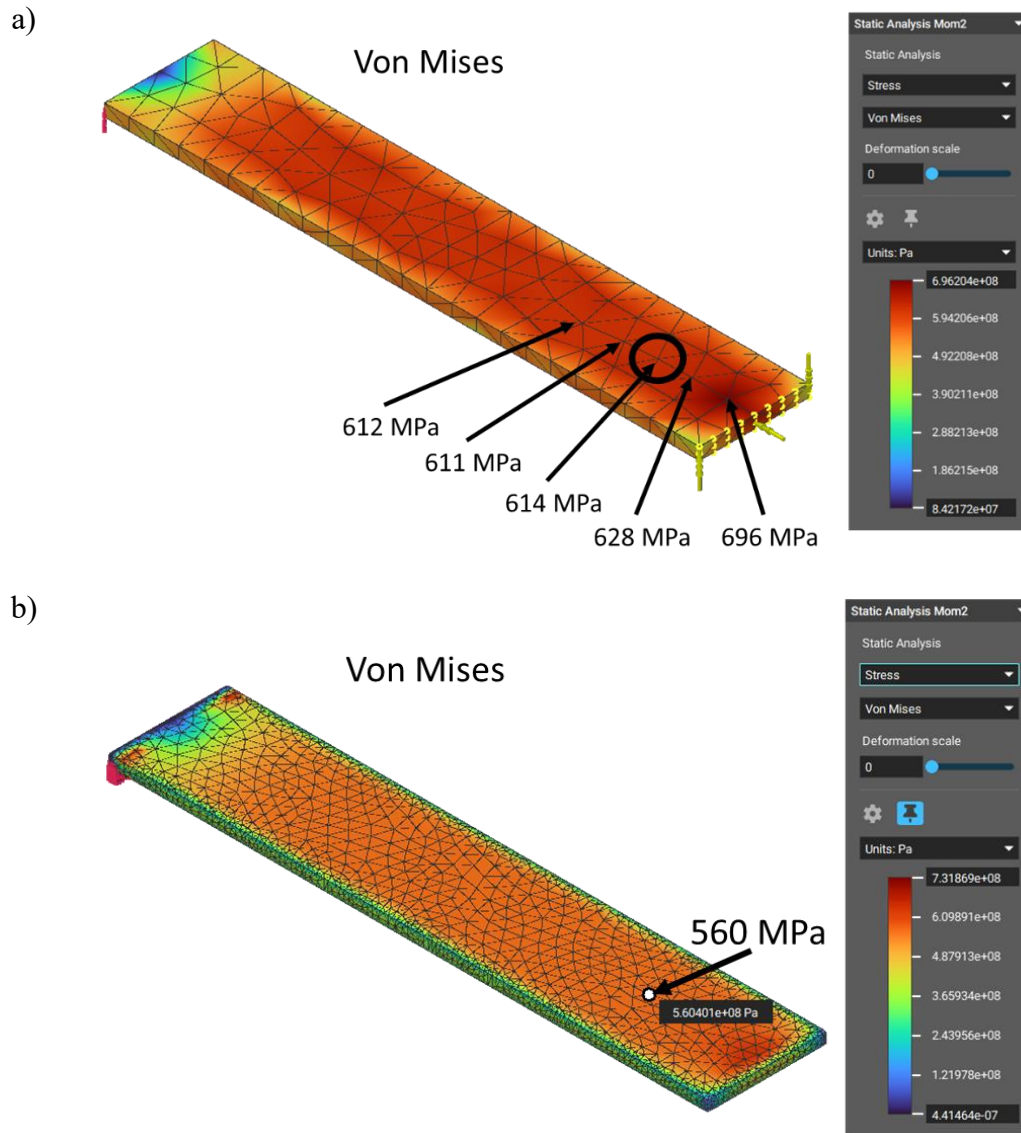


Figure 4.11. Finite element model von Mises stress plots of cantilever beam under torsion with a) coarse mesh, and b) refined mesh.

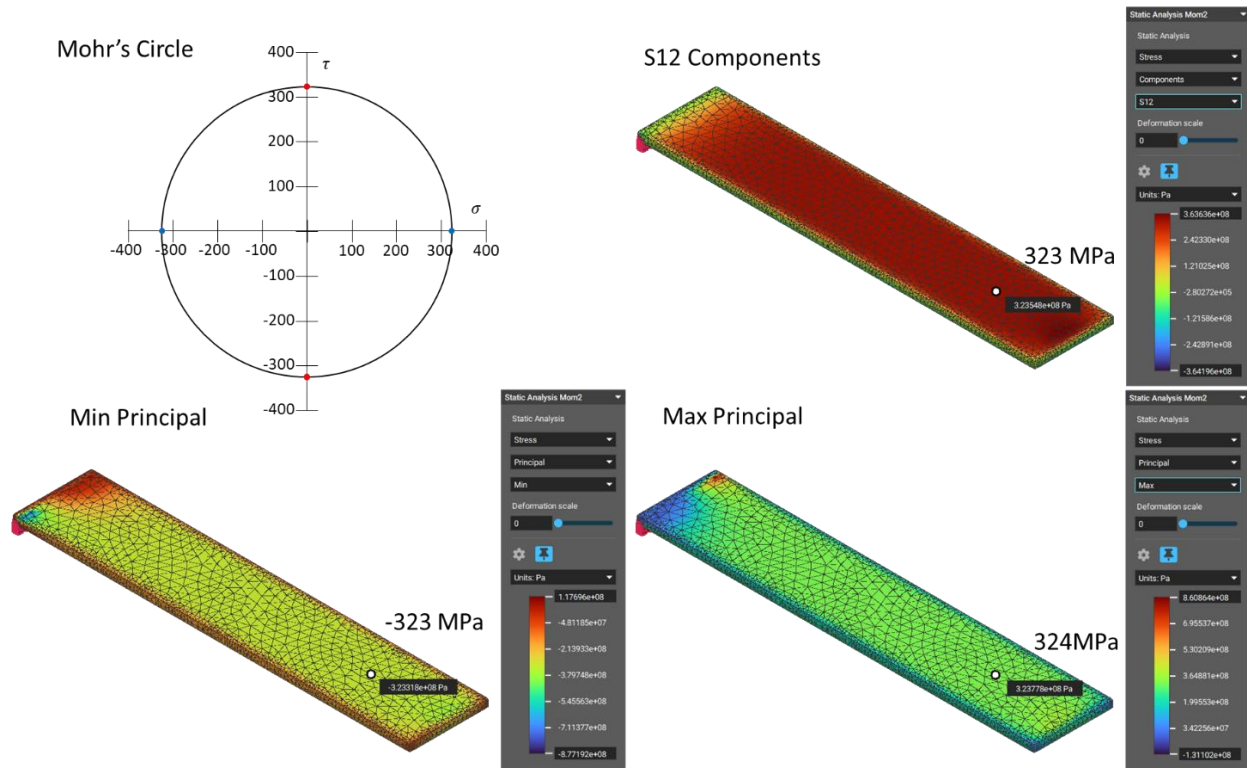


Figure 4.12. Cantilever beam refined mesh finite element model under torsion, with plots of Mohr's Circle, S12 Components Stress, Min Principal Stress and Max Principal Stress reflecting pure shear in the same regions as the von Mises stress.

This simple verification study yielded several important insights regarding the performance of the nTopology finite element solver. First, the numerical results demonstrated strong agreement with classical analytical solutions, with displacement and stress predictions falling within a 5% margin of error. Second, accurate correlation in both displacement and stress fields was achieved using quadratic tetrahedral elements, confirming their suitability for modeling complex geometries. Third, simulations involving bending and distributed loading conditions exhibited reliable behavior within the linear elastic regime.

While the torsional loading case showed good agreement in terms of angular displacement and twist, the stress values reported by the initial finite element model were significantly elevated.

That discrepancy was attributed to insufficient mesh density across the beam width, underscoring the sensitivity of stress predictions to element resolution near boundary conditions. Given that the lattice beam specimens in this study were specifically designed and optimized for flexural loading under 4-point bend, the verification results support the reliability of displacement predictions, even in regions where only a single element spans the thickness. However, caution must be exercised when interpreting stress results in areas subject to torsional deformation, as coarse meshing may artificially inflate stress magnitudes and imply premature yielding. Additionally, as is typical in finite element analysis, stress concentrations near boundary conditions should be interpreted with care, as they often reflect numerical artifacts rather than physical behavior.

4.5.2 *Solid Beam in 4-point Bend*

The second validation step for the nTopology FEA solver focused on assessing the accuracy of the simulation methodology employed for the 4-point bend configuration. Following the successful verification of the meshing strategy using a simplified cantilever beam model, the same meshing approach was applied to a solid beam with identical dimensions and geometry as the lattice beam test specimens. Given that the lattice core topology optimization was predicated on the 4-point bend load case, it was imperative to confirm the correctness of the applied boundary conditions and the implementation procedure. Failure to rigorously validate this setup would undermine the credibility of the optimization framework and compromise the reliability of the resulting optimization strategy outcomes.

To accomplish this check, the same 4-point bend simulation used for the lattice beam configurations was implemented for a solid bar of Ti-6Al-4V of similar size and shape as shown in Figure 4.13. In length, width, and height, the bar was modeled with the dimensions of 192 mm, by 48 mm, by 25.4 mm, respectively. It was meshed with 293,742 quadratic tetrahedron elements

and given the general material properties for Ti-6Al-4V: a Young's Modulus of 113.8 GPA, Poisson's Ratio of 0.342, and a density of 4.43 g/cm³. Two support boundary conditions (BC1 and BC2) were modeled on the underside of the beam, leaving a span of 142 mm. One boundary condition (BC1) was fixed displacement in X/Y/Z and allowed rotation only in Y, while the second boundary condition (BC2) allowed displacement in X and rotation in Y. Two 100 kN loads (F) were applied at the 1/3 and 2/3 span distance ($a = 0.0473$ m) between the supports along the upper surface as described in ASTM C393.

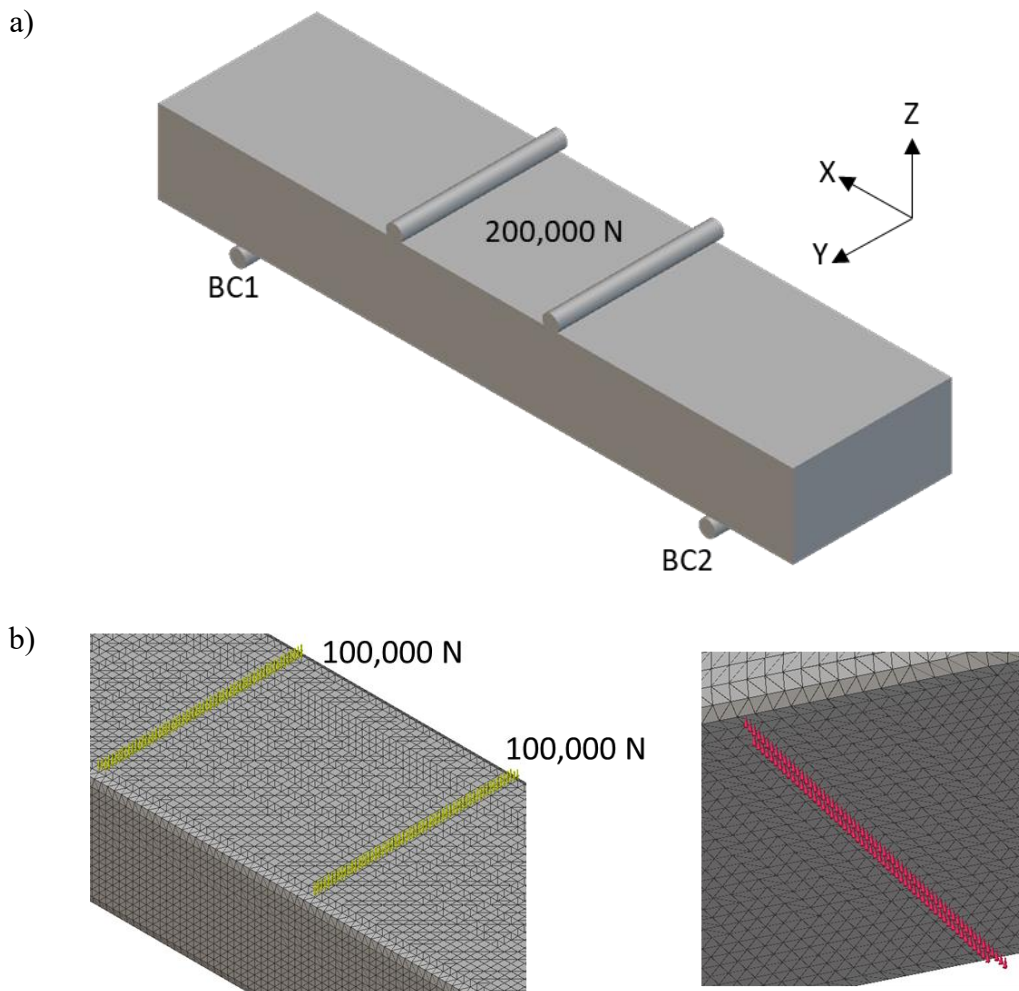


Figure 4.13. a) Isometric view of 4-point bend simulation model with 3rd span loading, b) Top view input load and bottom view support boundary conditions with respect to the FEM mesh.

Figure 4.14 displays the FEM displacement at the midpoint was $\delta_{max} = 1.49$ mm, while the displacement at the load application point was $\delta_a = 1.32$ mm. It also shows that the von Mises stress at the mid span lower surface to be 917 MPa with a strain of $8.09 \cdot 10^{-3}$. These FEA values for displacement, stress, and strain were compared in Table 4.2 with the analytical method from Section 4.4.2. Overall, good agreement was found between the hand calculations and the nTopology FEA simulation. While the error for the displacements was higher than desired at 9.5% and 11.9%, the magnitude of the differences was almost negligible at less than 0.2 mm. The error was much lower when comparing the von Mises stress and strain that resulted in 1.1% and 4.9% error, respectively. Thus, it was reasonable to claim that the nTopology finite element modelling method employed was valid for the prediction of displacement and stress with the 4-point bend simulations.

Table 4.2. Comparison of FEA stress results with hand calculations for the solid titanium beam in 4-point bend.

Load Case	4-point Bending			
Measure	δ_{max} (mm)	δ_a (mm)	σ (MPa)	ϵ (10^{-3})
Analytical	1.36	1.18	916	8.05
FEA Quad	1.49	1.32	917	8.09
error	9.5%	11.9%	1.1%	4.9%

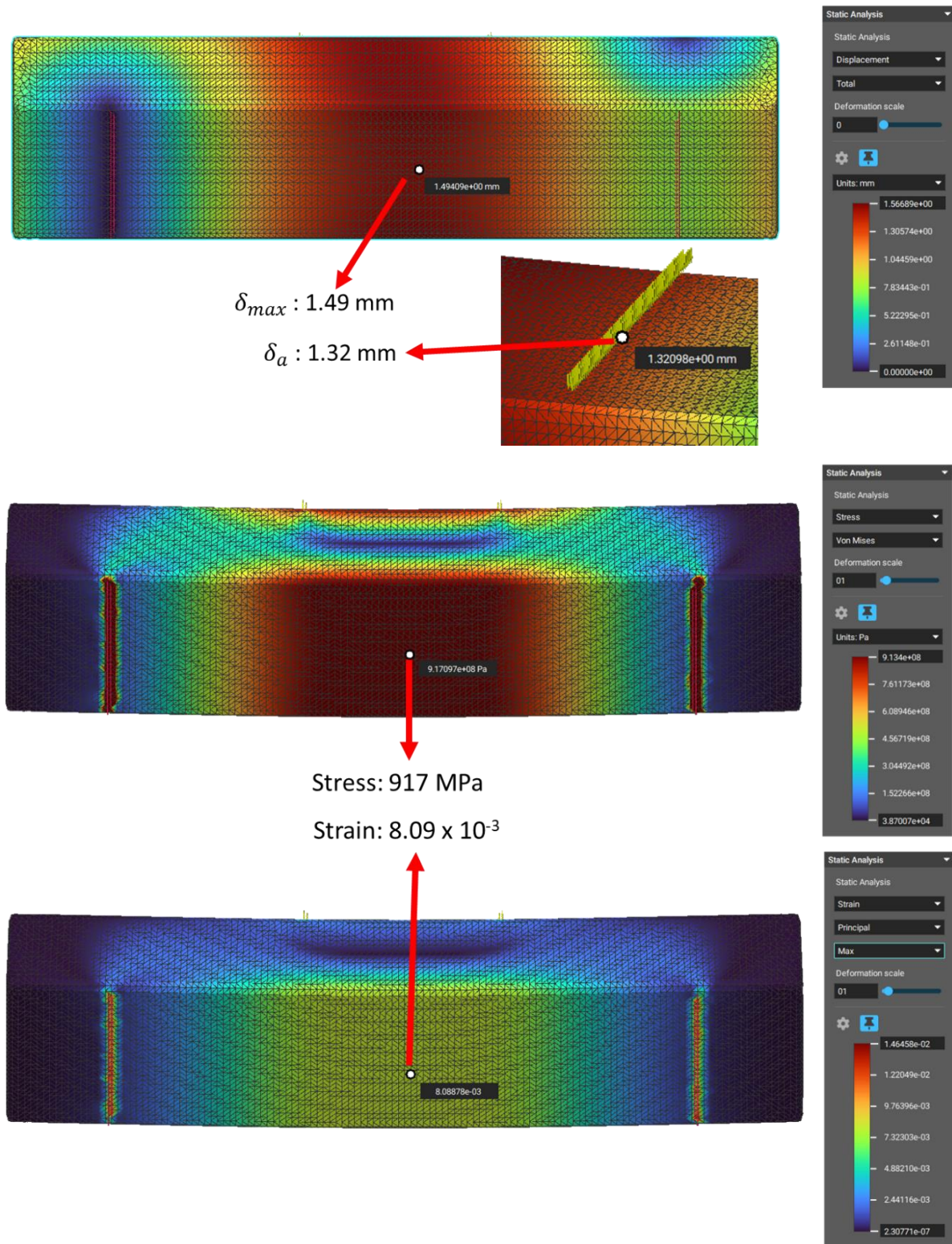


Figure 4.14. Displacement (top), von Mises stress (middle), and strain (bottom) plots of 4-point bend simulation model with 3rd span loading.

4.6 OPTICAL MICROSCOPY

Optical microscopy was employed as a primary characterization technique to examine the microstructural features and surface morphology of fabricated specimens. This method provides a means of visualizing material architecture at the microscale, enabling qualitative and quantitative assessment of features such as grain boundaries, phase distribution, and defect structures. The use of optical microscopy was particularly valuable in establishing baseline structural integrity alongside more advanced imaging methods. Sample preparation protocols, imaging parameters, and post-processing techniques were carefully selected to ensure reproducibility and clarity of the acquired images.

The performance and reliability of titanium components depends on the microstructure of the material. Ti-6Al-4V can go from fairly ductile to very brittle, with very little strain to failure, based on the quality of the feedstock material and processing conditions. As was thoroughly discussed in Section 2.3.3, a Grade 5 Ti-6Al-4V alloy is considered a dual phase ($\alpha + \beta$)-type. The microstructure produced by EB-PBF appears as a basket-weave (Widmanstätten) such that lamellar α phase surrounds wavy columnar prior β grains at the grain boundaries. Evaluation of the grain orientations plays a crucial role in understanding fracture behavior, and viewing the microstructure can help identify defects, inconsistencies, and deviations that provide insight into experimental observations. Therefore, understanding the relationship between microstructure and mechanical properties is fundamental to materials science, and a critical step in ensuring components meet quality and performance criteria.

To evaluate the microstructure of printed titanium samples, small sections were carefully cut using an abrasive saw with water as coolant. The sections were mounted in graphite-based mount media by means of a conventional mount press. Subsequent polishing of the mounted specimens

was accomplished with an EcoMet 30 semi-automatic polisher, using 400, 600, and 800 grit silicon carbide (SiC) paper. A rotation speed of 300 RPM was set for 3 minutes per grit, with a polishing load of 20 N/mount and water on. Polishing was continued using 1 mL DiaLube diamond suspension, with a particle size of ~ 3 μm , for 5-6 min. The final mirror-like finish was obtained via chemical attack polishing for 5 minutes. The chemical mixture consisted of 0.5 mL NH_4OH , with 2 drops of H_2O_2 , in 10mL colloidal silica that contained ~ 0.05 μm silica particle size. Following the polishing, the samples were etched for 8 seconds using Kroll's reagent, which consisted of 2% HF, 6% HNO_3 , and 92% H_2O . Once the microstructure had been revealed, optical images were obtained using an Olympus BX51 optical microscope with magnifications of 50x, 100x, 200x, 500x and 1000x.

For macroscopic imaging and dimensional analysis of manufactured specimens, a Keyence VR-3100 three-dimensional profile measurement microscope was utilized. This instrument employs high-intensity LED illumination in conjunction with a 4-megapixel monochrome CMOS sensor to capture one-shot fringe projection images. Fringe projection technology enables rapid, non-contact acquisition of geometric parameters, including length, height, volume, and angular measurements, across complex surface profiles. To enhance image clarity on specimens exhibiting significant topographical variation, the system incorporates image stacking techniques that synthesize multiple focal planes into a single composite image. For comprehensive visualization of the beam specimens, multiple fields of view were acquired at 40x magnification and subsequently stitched together using the instrument's proprietary software to generate high-resolution panoramic micrographs.

In summary, optical microscopy provided a robust platform for the microstructural characterization and macroscopic surface morphology for the various manufactured specimens.

The use of an Olympus BX51 optical microscope in combination with precision polishing and etching techniques revealed the complex EB-PBF Ti-6Al-4V microstructure in excellent clarity. The Keyence VR-3100 fringe projection and image stacking enabled enhanced visualization of complex lattice topographies, while stitched high-magnification views facilitated comprehensive inspection of beam failure. However, while optical methods are effective for capturing large-scale structural attributes, they are inherently limited in resolving fine surface details. To overcome these limitations and achieve higher-resolution imaging at the microscale, scanning electron microscopy (SEM) was employed as a complementary technique in the subsequent phase of analysis.

4.7 SCANNING ELECTRON MICROSCOPE

When looking to understand the failure behavior of a material, observation of the fracture is a critical step. The characteristics of the fracture surface can reveal the cause of failure, which may be due to overload, fatigue, corrosion, or other mechanisms. Identification of the root cause can help with avoiding failure in critical applications. A fracture surface can also provide insights into the failure mechanism, where a brittle fracture will appear different from a ductile fracture. For quality assurance, a fracture surface analysis can identify if a crack originated at a defect such as an inclusion or void, or if defects may have contributed to the speed of crack propagation. This type of analysis is best conducted under a scanning electron microscope (SEM) where high image resolution and magnification can be obtained without the lighting and focus challenges associated with optical microscopes.

Fracture surface characterization was conducted using a FEI XL30 Sirion scanning electron microscope. Owing to the limited chamber dimensions, specimens were sectioned into 20 mm cubic samples via precision cutting with an abrasive saw. For sandwich beam specimens, the

exposed cross-sectional surfaces were manually polished using sequential grades of silicon carbide (SiC) abrasive paper, 400, 600, and 800 grit, under continuous water lubrication to minimize thermal and mechanical damage. Post-polishing, samples underwent solvent cleaning with acetone followed by isopropyl alcohol, and were subsequently dried in a convection oven to eliminate residual moisture. The prepared specimens were mechanically secured to the SEM stage using a clamping mechanism. To facilitate imaging from multiple orientations, samples were manually rotated in 90-degree increments between scans. SEM imaging was employed to assess surface morphology, evaluate fracture characteristics, and identify potential microstructural defects or anomalies.

Chapter 5. RESULTS

The results of this work provide a comprehensive analysis of the microstructural and mechanical properties of the waste L-PBF titanium powder utilized as the feedstock for EB-PBF components. Detailed evaluations of the material were conducted for chemical composition, flowability, and morphology, before conducting experiments with an Arcam A2X machine. Initial trials required parameter adjustments to the preheat conditions before manufactured samples could be produced. Subsequent manufacturing quality assessments were carried out to determine what design parameters were critical to producing high-quality lattice structures for sandwich panel cores. These assessments also included the mechanical characterization of the as-provided material under tensile and compressive loading conditions.

Determination of the Ti-6Al-4V material properties became the inputs for the finite element simulation work that guided the design and topological optimization of the lattice sandwich specimens. Application of a SIMP based compliance minimization strategy and a von Mises stress minimization strategy resulted in novel architectures for the strut-type Octet and Kelvin lattices, as well as the TPMS-type Gyroid, Primitive, and Diamond lattices. In addition to a linear-static finite element analysis, a second area moment of inertia analysis provided further insight into the bending stiffness of these architectures. This research culminated in 4-point bend flexural experiments of the EB-PBF produced lattice sandwich structures, and the subsequent failure behavior analysis. Once all builds were completed, a final tensile and compressive analysis was performed to compare how the mechanical properties of the waste L-PBF powder had changed over the course of the experiments.

The findings revealed significant variations in the flexural behavior of the different lattice geometries. In each case, the topology optimization strategies demonstrated improvement over a

uniform core geometry, while the TPMS-type lattices showed superior performance over the strut-types. The variety of core geometries used in this work was sufficient to successfully exhibit several different failure modes. Those findings offer valuable insights to engineers, such that they could select the core type that is most advantageous to their application. The subsequent sections present a thorough description of the results of this research, and they highlight key trends relevant to the use of EB-PBF titanium in high-performance engineering contexts.

5.1 FEEDSTOCK POWDER ANALYSIS

A comprehensive evaluation of the powder feedstock is essential to ensure consistency, reliability, and performance in additive manufacturing, as well as other powder metallurgy applications. This section presents a multi-faceted characterization of the powder materials, encompassing chemical composition, flowability, and morphological attributes. Chemical composition analysis was conducted to verify elemental purity and detect potential contaminants that may influence phase formation, mechanical properties, or process stability. Techniques such as energy-dispersive X-ray spectroscopy (EDS) and inductively coupled plasma optical emission spectroscopy (ICP-OES) were employed to quantify the elemental constituents with high precision.

Flowability assessments were performed to determine the powder's suitability for uniform layer deposition and consistent material distribution during processing. A valuable metric for the flow dynamics under gravity-driven conditions is Hall flow rate. Morphological analysis was carried out using scanning electron microscopy (SEM) to examine particle shape and size distribution. These parameters critically affect powder packing density, melt pool behavior, and final part integrity. Particular attention was given to the presence of satellite particles, agglomerates, and irregular geometries, which can adversely impact processability and part quality. Together, these analyses provided a complete awareness of the unknown feedstock

physical and chemical characteristics, enabling informed decisions in process optimization and quality control. Additional information supporting the feedstock analysis results presented can be found in Appendix A.

5.1.1 Chemical Composition

Three representative samples of the as-received waste L-PBF Ti-6Al-4V powder were extracted from the bulk material and submitted for chemical composition analysis. The results, summarized in Table 5.3, display a titanium concentration consistent with nominal alloy specifications, a slightly reduced aluminum content, and a marginally elevated vanadium level. Of particular importance was the oxygen concentration, given its critical influence on the mechanical properties and classification of titanium alloys. The measured oxygen content was low and approached the threshold for Grade 23 Extra Low Interstitial (ELI) specification ($\leq 0.13\%$), indicating minimal oxidation during prior handling and storage.

Table 5.3. Chemical analysis of as-received L-PBF Ti-6Al-4V powder feedstock in comparison to a standard Ti-6Al-4V disk, manufacturer provided Ti-6Al-4V, and ASTM F3001 chemical composition standard

	Ti (%)	Al (%)	V (%)	Fe (%)	Zr (%)	Cr (%)	Si (%)	Sn (%)	O (%)	N (%)	H (%)	C (%)
L-PBF Run 1	88.6	5.0	6.0	0.14	0.09	0.00	0.03	0.02	0.138	0.315	0.008	0.013
L-PBF Run 2	90.5	5.1	4.1	0.14	0.11	0.09	0.03	0.02	0.141	0.318	0.008	0.013
L-PBF Run 3	90.2	5.1	4.4	0.14	0.12	0.08	0.03	0.02	0.139	0.319	0.008	0.012
L-PBF avg	89.8	5.1	4.8	0.14	0.11	0.06	0.03	0.02	0.139	0.317	0.008	0.013
Std SRM 654A Run	89.4	6.4	3.4	0.18	0.08	0.22	0.19					
Std SRM 654A Cert	89.4	6.3	3.9	<0.2	0.00	0.00	0.00					
Arcam Ti-6Al-4V	Bal.	6.0	4.0	0.1	-	-	-	-	0.1	0.01	<0.003	-
ASTM F3001	Bal.	5.5- 6.5	3.5- 4.5	<0.25	-	-	-	-	<0.13	<0.05	<0.012	<0.08

Notably, the nitrogen concentration exceeded the expected limit ($\leq 0.03\%$) by approximately an order of magnitude, which may warrant further investigation into powder exposure conditions

or atmospheric contamination during prior use. Despite this anomaly, the overall chemical profile did not exhibit any critical deviations from acceptable ranges, and the powder was deemed chemically suitable for use in the EB-PBF experimental trials as Ti-6Al-4V alloy.

5.1.2 *Flowability*

Flowability assessments were conducted in accordance with ASTM B213 using a calibrated Hall Flowmeter. The as-received powder exhibited a Hall flow rate of $FR_H = 27$ s/50 g per method 1. Following four cycles of processing through the Powder Recovery System (PRS) and subsequent sieving, the flow rate improved to $FR_H = 24$ s/50g per method 1, indicating enhanced flowability. Reference data from AP&C, a recognized supplier of metal powders for additive manufacturing, reports typical flow rates of 28 seconds for powders within the 15–45 μm size range (commonly used in L-PBF applications) and 25 seconds for powders in the 45–106 μm range (typical for EB-PBF processes). The observed reduction in flow time post-processing is therefore consistent with a shift in particle size distribution toward coarser fractions, which are known to exhibit superior flow characteristics due to reduced interparticle friction and improved packing behavior.

5.1.3 *Morphology*

Powder morphology was characterized through image analysis performed in MATLAB, utilizing high-resolution micrographs obtained via a Scanning Electron Microscope (SEM). This instrument employs low-kV secondary electron imaging with through-the-lens detection, enabling detailed visualization of surface features and particle geometry. Three distinct powder conditions were examined: 1) the as-received state, 2) post-processing via the Powder Recovery System (PRS) and sieving, and 3) the fine particle fraction isolated by the PRS. Representative SEM imagery of the as-received powder is presented in Figure 5.1a, illustrating a heterogeneous

distribution of particle sizes and morphologies. In contrast, Figure 5.1b depicts the fines fraction, which exhibits a more uniform and spherical particle population, consistent with the expected characteristics of smaller particles selectively removed from the bulk feedstock during PRS operation.

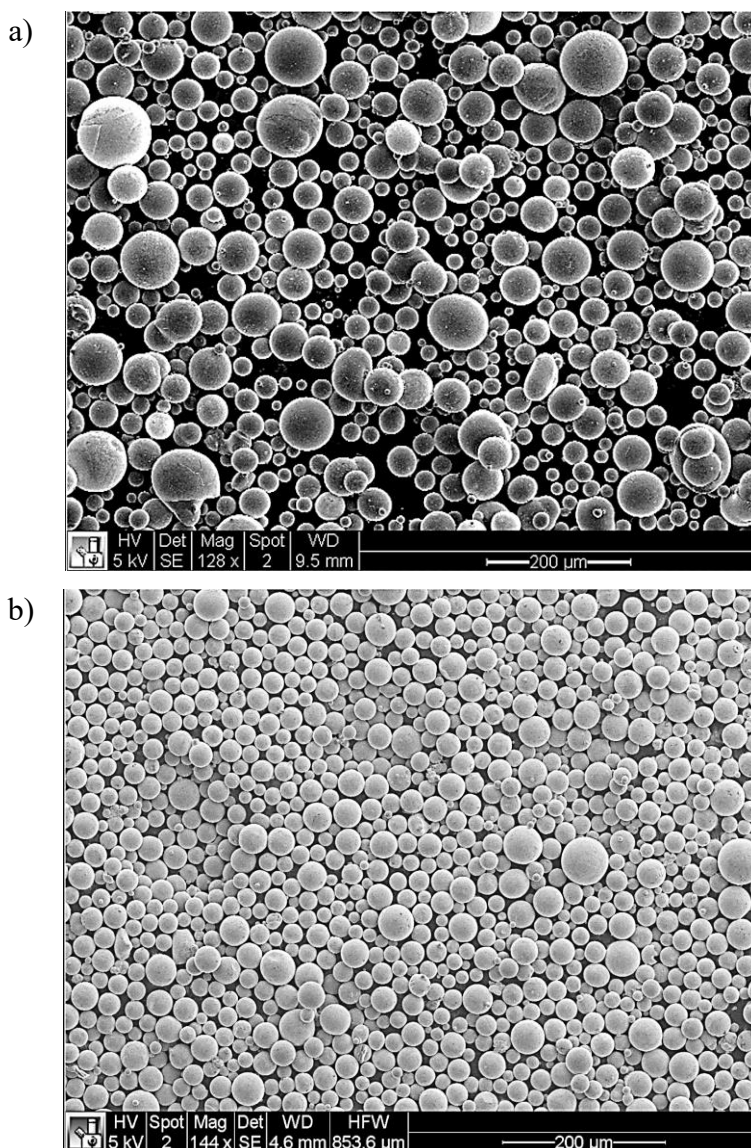


Figure 5.1. SEM image of waste L-PBF Ti-6Al-4V powder in the a) as-received condition, and b) fines removed by the PRS (WD: working distance, DET: detection scanning electron, HV: voltage, Spot: beam current)

The particle size distribution (PSD) analysis, based on measured diameters, is summarized in Table 5.4. The as-received powder exhibited a mean particle diameter of 27.53 μm . Following processing through the PRS, sufficient fine particles were removed to increase the average particle size to 33.04 μm . Notably, the lower decile (D10) shifted from 16.69 μm to 20.23 μm , indicating a reduction in the proportion of ultrafine particles. Conversely, the upper decile (D90) decreased from 56.89 μm to 50.68 μm , displaying a modest reduction in the largest particle fraction, potentially due to sieving large contaminants or agglomerate breakdown.

Table 5.4. Particle size distribution of diameters in the 10/50/90 percentiles

	10th Percentile (μm)	50th Percentile (μm)	90th Percentile (μm)	Sample Size (count)
As Received	16.69	27.53	56.89	4913
Sieved	20.23	33.04	50.68	4026
Fines	13.34	21.23	30.1	3563

Quantitative image analysis was performed using a custom MATLAB script, which identified and counted circular features corresponding to individual particles. The sample sizes analyzed identified 4,913 particles in the as-received condition, 4,026 particles post-sieving, and 3,563 particles from the fines fraction. Despite efforts to increase the overall PSD of the recycled L-PBF powder, the resulting distribution remained significantly finer than the typical particle size range employed in EB-PBF applications, which generally favors coarser feedstock for optimal process stability and flowability.

Although the D10, D50, and D90 metrics presented in Table 5.4 suggest only a modest shift in PSD following PRS processing, the histogram plots in Figure 5.2 offer a more nuanced visualization of the distributional changes. As illustrated in Figure 5.2a, the PSD of the as-received powder exhibits a left-skewed profile, characterized by a substantial proportion of particles falling below the median (D50) diameter. The calculated D50 value is elevated relative to the modal

particle size due to the presence of a non-negligible population of outlier particles exceeding 50 μm in diameter, which disproportionately influence the central tendency.

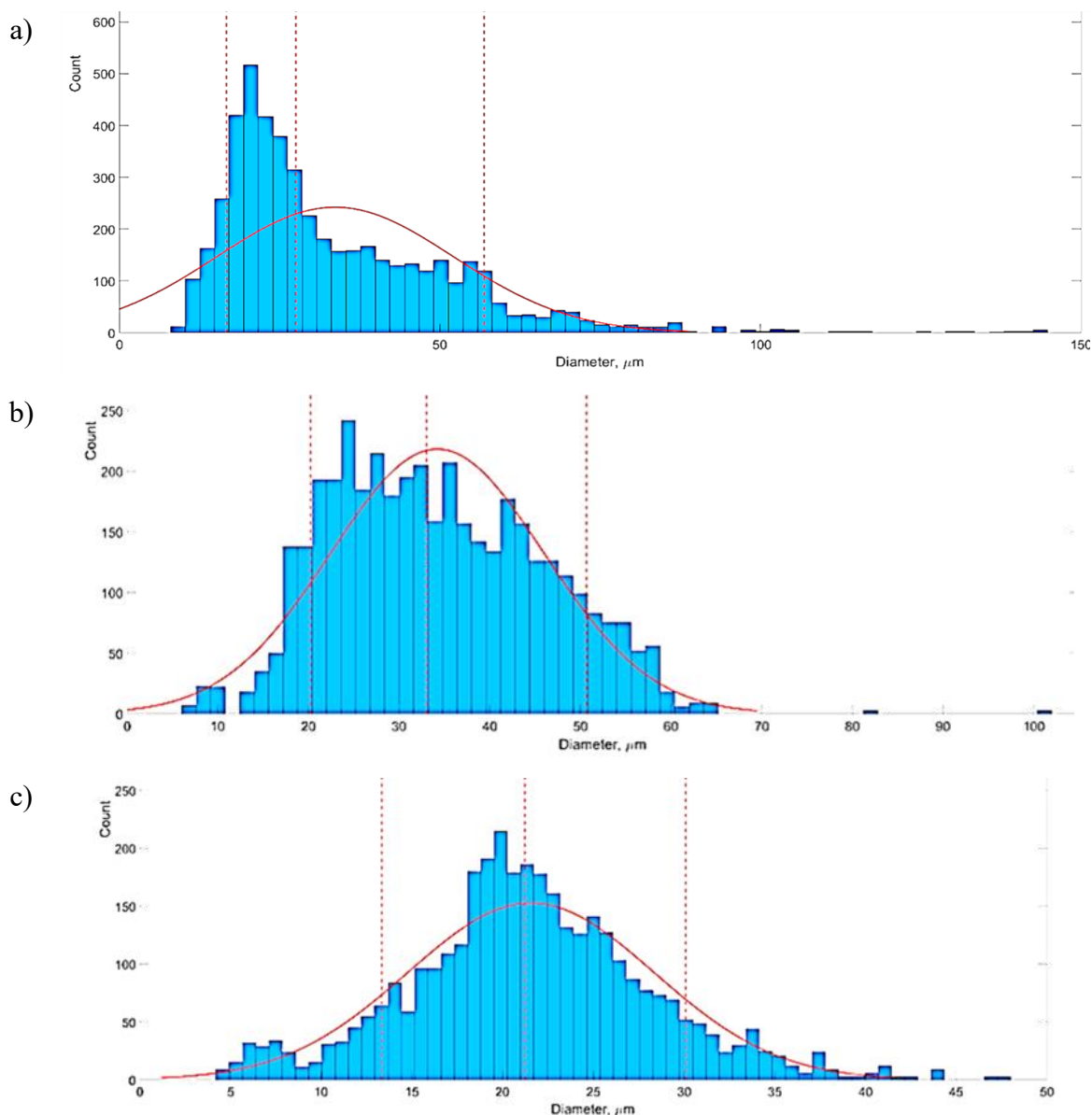


Figure 5.2. Particle size distributions for the waste L-PBF feedstock, a) as-received, b) sieved, and c) fines removed, with vertical lines indicating the D10/50/90 values.

Post-processing via the PRS and sieving resulted in a particle diameter distribution, as shown in Figure 5.2b, that more closely approximates a normal distribution. The D10, D50, and D90

values became more evenly spaced, indicating a more symmetric spread of particle sizes. The reduction in the D90 value is attributed to the effective removal of larger particles ($>50\text{ }\mu\text{m}$), which were selectively filtered out during processing. Figure 5.2c illustrates the PSD of the fines fraction recovered by the PRS, highlighting its capability to isolate smaller particles. This subset also exhibits a near-normal distribution, with a narrow size range predominantly between $5\text{ }\mu\text{m}$ and $45\text{ }\mu\text{m}$, further confirming the efficiency of the PRS in segregating fine particulates from the bulk feedstock.

5.2 EB-PBF EXPERIMENTAL TRIALS

This section presents the experimental results evaluating the viability of waste L-PBF Ti-6Al-4V powder for use in the Electron Beam Powder Bed Fusion (EB-PBF) process, as well as the outcomes of preliminary manufacturing quality assessments conducted prior to the fabrication of sandwich beam specimens. Following comprehensive powder characterization confirming chemical, flowability, and morphological suitability, the recycled feedstock was loaded into the Arcam A2X EB-PBF system. To establish baseline process compatibility, an initial build that was previously validated using EB-PBF powder supplied by the machine manufacturer, was replicated using the recycled material.

Early challenges related to powder instability, specifically smoking during the preheating phase, were mitigated through iterative adjustments to the build plate preheat parameters. The development and implementation of a novel preheating protocol successfully stabilized the powder bed, enabling consistent layer deposition and thermal control. This advancement facilitated the reliable execution of subsequent builds, including those for manufacturing quality evaluation, mechanical testing, and final sandwich beam fabrication.

5.2.1 Line Build

The initial batch of filtered waste L-PBF Ti-6Al-4V powder was introduced into the feed hoppers of the Arcam A2X EB-PBF system. Default manufacturer-supplied process parameter themes for melting and wafer scanning were retained, while the modified build plate preheating protocol to mitigate powder instability was implemented. The selected build configuration replicated a previously validated setup using standard EB-PBF powder, thereby isolating the powder source as the sole experimental variable.

This trial, referred to as the Line Build [34], was designed to evaluate melt pool topology by producing 4 mm long single-beam-width melt lines, one layer thick, atop the powder bed. Upon completion of the revised preheat procedure, the extended preheating cycle required 1 hour and 39 minutes to reach the target temperature. The build proceeded without incident, and no smoke events were observed, confirming the stability of the recycled powder under the revised thermal conditions. Figure 5.3 presents the resulting melt line morphology in comparison to a prior build conducted with standard EB-PBF Ti-6Al-4V powder.

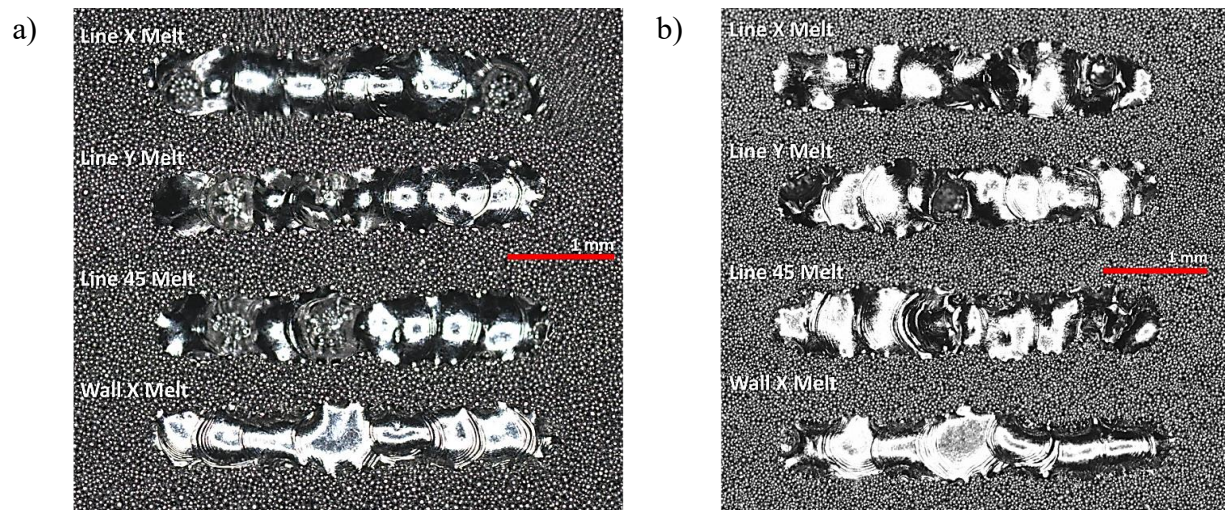


Figure 5.3. Line Build experiment results using (a) manufacturer provided EB-PBF Ti-6Al-4V powder, and (b) non-standard waste L-PBF Ti-6Al-4V powder.

Microscopic examination of the melt lines revealed a sequence of overlapping, circular melt pools embedded within the powder bed. In certain regions, partially melted pockets were observed, containing residual unmelted powder likely introduced post-solidification due to powder bed redistribution following beam termination. Peripheral to each melt pool, raised ridges resembling solidified waveforms or ripple-like structures were evident, indicative of localized thermal oscillations during solidification.

Although precise quantification of individual melt spots was challenging due to overlap, an estimated eight discrete melt pools, each approximately 0.6 mm in diameter, were discernible. The upper surface of the melt region exhibited a smooth, reflective finish, while the overall melt geometry was concave, penetrating approximately 0.15 mm into the powder bed [34]. Notably, the interface between the melt zone and the surrounding sintered powder was characterized by pronounced surface irregularities and chaotic morphology. This observation underscores a contributing factor to the inherently poor surface finish of EB-PBF components, particularly in regions formed via contour scanning strategies that employ discrete spot melting techniques.

Between the two powder feedstocks presented in Figure 5.3a and Figure 5.3b, there were clear observable differences. The surrounding particulate morphology showed that the EB-PBF powder exhibits a coarser PSD relative to the processed waste L-PBF powder. Furthermore, the melt pool topology associated with the waste L-PBF powder demonstrated enhanced smoothness, uniformity, and overall surface integrity, alluding to superior melting behavior under the applied EB-PBF conditions. The waste L-PBF powder produced fewer discernible solidification ripples, and the presence of unmelted pockets was both reduced in size and less well-defined compared to the EB-PBF counterpart. These outcomes, coupled with the successful execution of the build

without smoking, substantiated the viability of utilizing the non-standard, waste L-PBF powder within the EB-PBF process for the scope of this research.

5.2.2 *Preheat Parameter Adjustment*

The experimental setup for the initial lattice build (designated B2) mirrored that of the previously successful Line Build (B1), with the exception of an increased ramp rate for the build plate preheat process. The initial beam current was elevated to 10 mA, compared to 1 mA in prior trials, and incrementally increased at lower temperature thresholds than previously attempted. This modified approach attained the nominal beam current of 48 mA at the designated process initiation temperature of 730 °C within 43 minutes, corresponding to an average ramp rate of 16.4 °C/min, and was achieved without triggering any smoke events during the preheat phase.

Despite the successful preheat procedure, powder instability was observed during the Preheat 1 phase, specifically when the electron beam first interacted with the powder following the distribution of the initial layer. This interaction resulted in a smoke event; however, upon beam reactivation, the remainder of the layer was completed without further incident. The second layer proceeded without smoke, whereas the third layer exhibited renewed instability. A total of 17 intermittent smoke events were recorded across the first 33 layers, after which the build appeared to stabilize temporarily. Nevertheless, at layer 106, corresponding to a build height of approximately 5.3 mm, multiple suspected smoke events were inferred from arc trip warnings logged by the system, culminating in catastrophic build failure.

It was observed during the early layers of the build that the temperature dropped quickly. This was confirmed by the temperature profile from the machine log displayed in Figure 5.4a. In comparison to the successful Line Build in Figure 5.4b, there was a steep drop in temperature after initially reaching 730 °C. This drop was caused by the numerous smoke events and beam shut

downs that occurred in the first 33 layers. The Arcam software tries to maintain constant energy on every layer given the beam energy input and the energy lost to radiation and conduction; however, the software does not appear to compensate for any layer interruptions. Therefore, during the brief times that the beam would turn off and restart, excessive cooling was occurring.

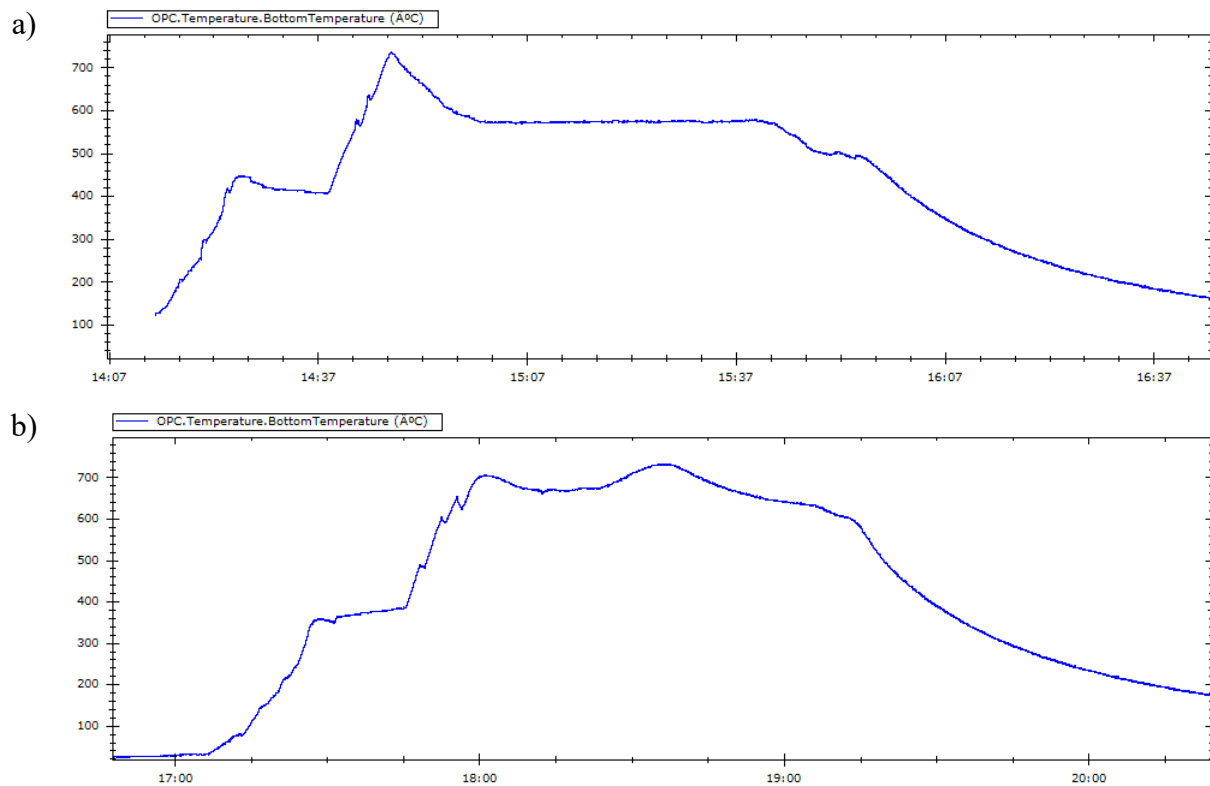


Figure 5.4. Arcam A2X temperature log from (a) failed initial Lattice Cube Build (B2), and (b) successful Line Build (B1). Note: time scales are not the same.

Although the precise cause of the final build failure remains undetermined, it is postulated that the early smoke events contributed to a reduction in chamber temperature, thereby destabilizing subsequent layers. One prevailing hypothesis attributed these initial smoke events to the accelerated preheat ramp rate, which may have insufficiently allowed the chamber environment to reach thermal equilibrium. While the powder surrounding the build plate gradually approached

the target temperature, the newly introduced powder from the feed hoppers, having not undergone adequate thermal conditioning, remained comparatively cooler and exhibited instability when exposed to the Preheat 1 beam current of 30 mA. This thermal mismatch likely triggered localized smoke generation upon beam interaction. In light of these observations, it was concluded that subsequent experimental trials utilizing the non-standard L-PBF powder would adopt the more conservative temperature ramp rate of approximately 7 °C/min, consistent with the protocol established during the successful Line Build, to promote uniform heat distribution and mitigate powder-induced instability.

The subsequent build (B3) successfully achieved the target build plate preheat temperature of 730 °C after 1 hour and 32 minutes, without exhibiting any signs of powder instability. The build proceeded to a height of 12.9 mm (6.9 mm into the specimens) prior to encountering an unanticipated filament failure, which terminated the process. At this juncture, the remaining powder inventory from the initial batch had decreased to 40.82 kg, raising concerns regarding its sufficiency to complete the intended build height. To address this potential shortfall, an additional 40.82 kg of powder was sourced from the original as-received lot. Following three processing cycles through the PRS and subsequent sieving, 33.57 kg of usable powder was recovered and integrated with the initial batch, resulting in a revised Batch 2 total of 74.39 kg.

The fourth experimental build (B4) employed the same build plate preheat temperature ramp rate protocol as utilized in builds B1 and B3, successfully attaining the target temperature of 730 °C within 1 hour and 30 minutes. Retrospective analysis of the preheat profiles indicated that the initial beam current increment of 1 mA did not contribute meaningfully to the thermal ramp and was therefore deemed superfluous. Consequently, this step was omitted in the next build (B5), while preserving the overall temperature ramp rate of approximately 7 °C/min to ensure thermal

consistency. A comparative plot illustrating the build plate temperature ramp profiles for the first five experimental builds is presented in Figure 5.5 (supporting data is located in Appendix C), highlighting the evolution of preheat strategies across the test series.

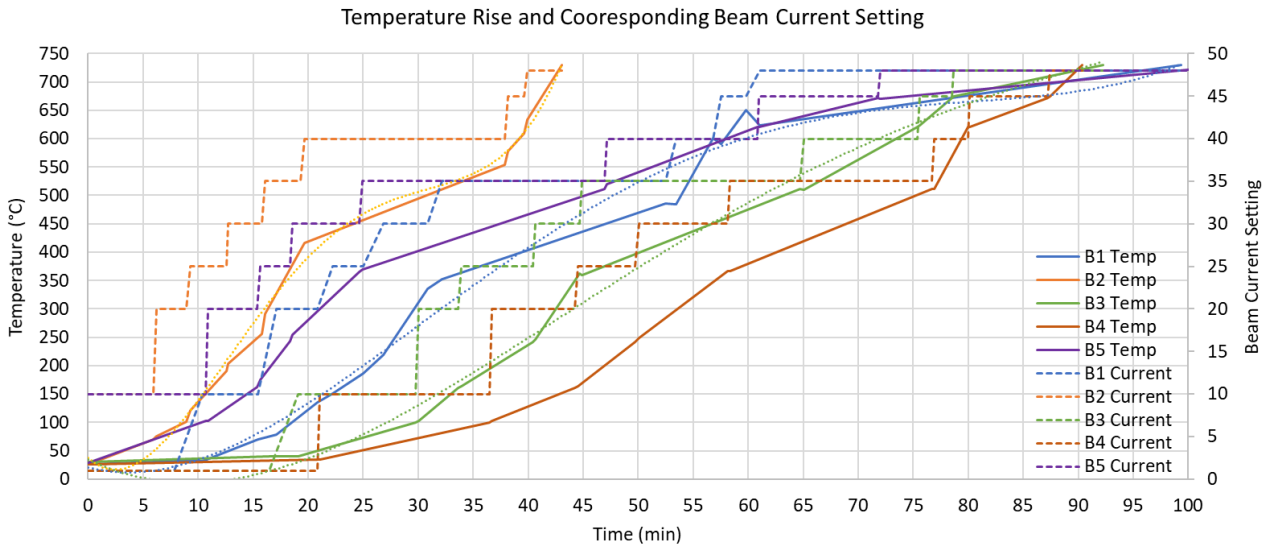


Figure 5.5. Temperature ramp rates and beam current increases for experimental builds 1-5 (B1-B5).

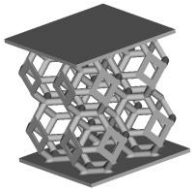
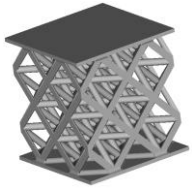
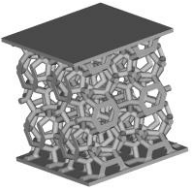
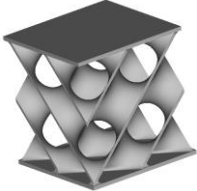
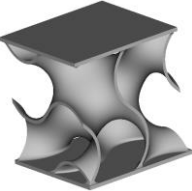
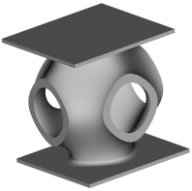
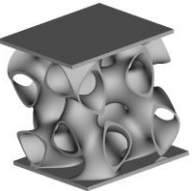
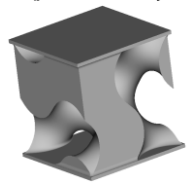
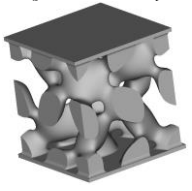
5.2.3 Lattice Manufacturing

Manufacturing evaluations of lattice architectures commenced with the design of cubic specimens measuring 25.4 mm (1 inch) per side, incorporating 0.8128 mm thick face sheets that corresponded to standard 0.032 inch sheet stock, and an internal lattice core. Due to limitations in explicit modeling capabilities within conventional CAD platforms, all 3D geometries were generated using implicit modeling via nTopology, leveraging its library of preprogrammed lattice configurations. The initial investigation concentrated on three principal lattice categories: 1) conventional strut-based lattices, 2) TPMS sheet structures, and 3) TPMS skeletal configurations. Representative strut-based lattices included the Kelvin cell, Octet truss, and Weaire–Phelan structure. TPMS

sheet configurations encompassed the Diamond, Gyroid, Primitive, and SplitP surfaces, while skeletal variants comprised the Diamond, Gyroid, and Lidinoid forms.

A total of ten lattice configurations, as detailed in Table 5.5, were selected to span a diverse range of topologies and to qualitatively evaluate powder removal feasibility. Geometries exhibiting occluded regions or enclosed volumes that could impede powder evacuation were excluded from consideration. TPMS sheet structures were modeled with a single unit cell and a uniform wall thickness of 1 mm. Strut-based lattices were defined with two unit cells along each principal axis and a strut diameter of 1.5 mm. Skeletal TPMS configurations were also based on a single unit cell; however, wall thickness was not applicable, as these structures were inherently designed to occupy approximately 50% of the volume with solid material.

Table 5.5. Sandwich sample configurations and the lattice relative density (ρ).

Strut				
	Kelvin ($\rho = 0.09$)	Octet ($\rho = 0.17$)	Weaire-Phelan ($\rho = 0.14$)	
TPMS sheet				
	Diamond ($\rho = 0.09$)	Gyroid ($\rho = 0.08$)	Primitive ($\rho = 0.14$)	SplitP ($\rho = 0.11$)
TPMS Skeletal				
	Diamond ($\rho = 0.50$)	Gyroid ($\rho = 0.50$)	Lidinoid ($\rho = 0.38$)	

The lattice cube specimens were arranged according to the build layout illustrated in Figure 5.6, using Materialise Magics for part placement and support generation. The build configuration also included nine tensile specimens conforming to ASTM E466 standards, featuring a nominal gauge diameter of 9 mm, along with additional components designated for other experimental investigations. All parts were positioned 5 mm above the build plate and supported by custom-designed supports spanning the vertical offset to ensure thermal conduction and mechanical stability during fabrication. Despite completing the custom chamber preheat cycle, Build B4, was only partially successful. While the lattice cube specimens were completed without incident, the build failed during the fabrication of the tensile specimens.

The hypothesized cause of failure was attributed to a significant reduction in melt area per layer following the completion of the lattice cubes. This abrupt transition resulted in localized heat accumulation due to shorter layer times and concentrated energy input over a smaller region of the powder bed. The excessive thermal buildup led to material swelling, which caused physical interference between the solidified parts and the rake mechanism. Contact between the swollen components and the rake teeth resulted in damage to the recoating system, leading to inconsistent powder distribution across subsequent layers. This issue progressively worsened, ultimately deforming the midsections of the tensile specimens to the extent that the rake could no longer transit the build area. The machine responded by initiating an automatic shutdown, terminating the build.

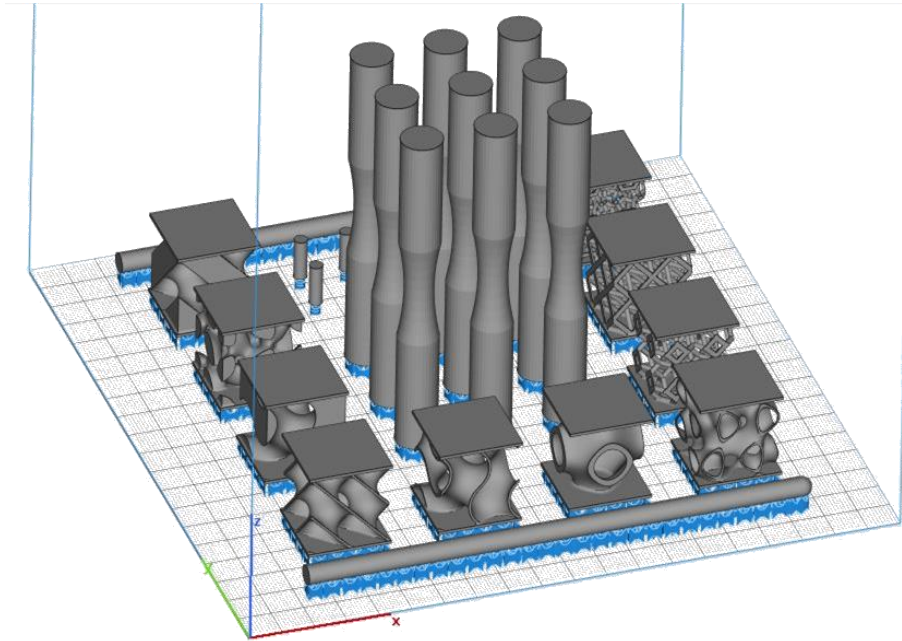


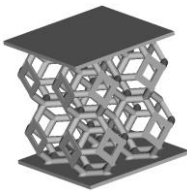
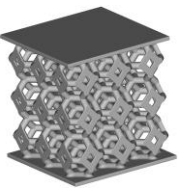
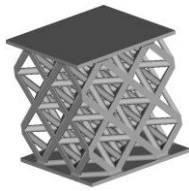
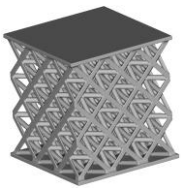
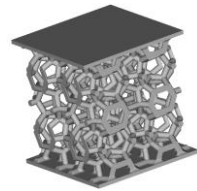
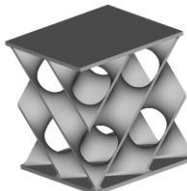
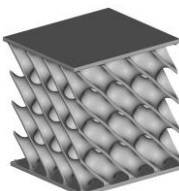
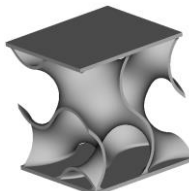
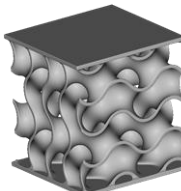
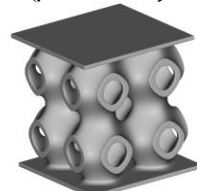
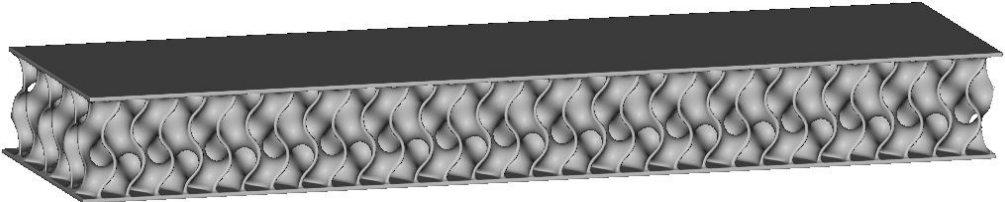
Figure 5.6. Sandwich lattice manufacturing quality build layout, version 1.

Despite the failure of the tensile specimens, the lattice sandwich samples were successfully fabricated and suitable for manufacturing quality assessment. To obtain complete tensile coupons for material behavior characterization, the build layout was restructured. Rather than positioning all lattice samples directly on the build plate, they were strategically distributed along the vertical axis of the build volume. This nesting approach was implemented to promote a more uniform melt area per layer, thereby mitigating thermal gradients and enhancing process stability. Additional modifications to the build configuration included the introduction of alternative lattice geometries and a reorientation of all samples by 90 degrees, aligning the face sheets parallel to the build direction.

This revised layout incorporated repeated instances of the Kelvin, Octet, Weaire–Phelan, Diamond, and Gyroid structures to facilitate comparative analysis of build orientation effects on structural integrity and surface quality. Furthermore, variations in lattice density were introduced: the Kelvin and Octet structures were expanded from 2^3 to 3^3 unit cells, while the Diamond, Gyroid,

and Schwarz Primitive structures were increased from 1 to 2^3 unit cells. The TPMS Skeletal-type structures were excluded from this iteration and replaced with a sandwich beam specimen designed in accordance with ASTM C393 standards. The beam featured a length of 190 mm, a width of 38.1 mm, and a height of 19.5 mm, with the unit cell dimensions configured to accommodate three cells across the width. Table 5.6 summarizes the lattice sandwich specimens included in the revised build configuration, which is illustrated in Figure 5.7.

Table 5.6. Sandwich sample configurations and the lattice relative density (ρ).

Strut					
	Kelvin ($\rho = 0.09$)	Kelvin ($\rho = 0.18$)	Octet ($\rho = 0.17$)	Octet ($\rho = 0.19$)	Weaire-Phelan ($\rho = 0.14$)
TPMS sheet					
	Diamond ($\rho = 0.09$)	Diamond ($\rho = 0.19$)	Gyroid ($\rho = 0.08$)	Gyroid ($\rho = 0.16$)	Primitive ($\rho = 0.20$)
TPMS Beam					
	Gyroid ($\rho = 0.16$)				

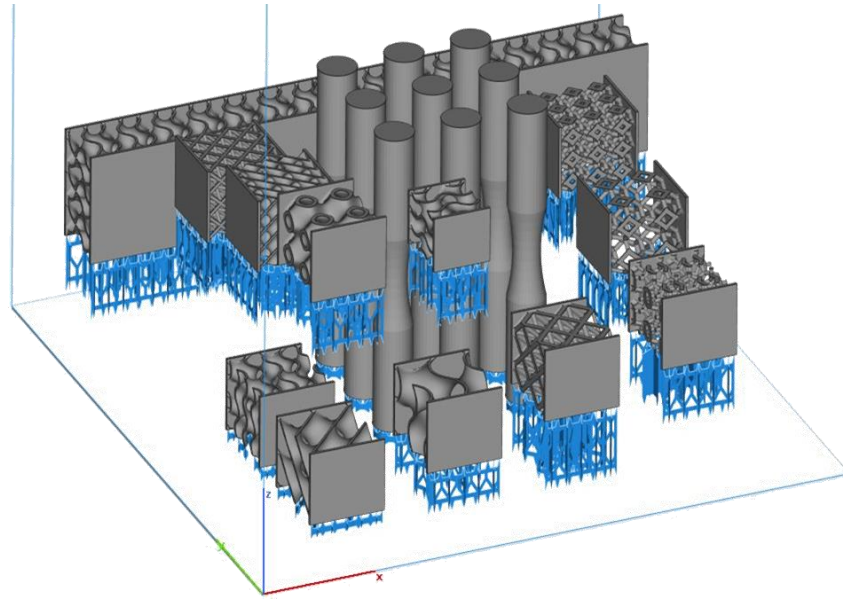


Figure 5.7. Sandwich lattice manufacturing quality build layout, version 2, with rotated and vertically nested samples.

The revised lattice cube build (B5) completed successfully, reaching the full intended build height without any thermal anomalies such as localized overheating or smoke generation. This outcome yielded a sufficient quantity of lattice sandwich specimens for comprehensive manufacturing quality evaluation, as well as tensile coupons suitable for subsequent material behavior characterization. The results of this build confirmed that no additional lattice cubes or beam specimens were necessary to finalize the design parameters for the concluding phase of the sandwich beam investigation. With the requisite specimens fabricated, the focus shifted to a detailed Manufacturing Quality Assessment, wherein the dimensional accuracy, surface integrity, and structural fidelity of the lattice sandwich samples were qualitatively analyzed.

5.2.4 *Manufacturing Quality Assessment*

To identify the most suitable lattice core configurations for four-point bending flexural testing of sandwich beam specimens, a total of 20 cube samples and one beam specimen were fabricated, as

outlined in Section 5.2.3. Each specimen underwent a qualitative evaluation across four key criteria: manufacturability, demonstration of AM capabilities, structural characteristics, and relevance to existing literature.

Manufacturability was assessed based on dimensional accuracy, resolution of overhanging features, absence of visible defects, and the complexity of post-processing, specifically powder removal using the PRS. In terms of demonstrating AM potential, lattice configurations were favored if they exhibited geometries that were infeasible through conventional manufacturing methods, featured intricate organic-like morphologies, or offered prospects for multifunctional performance. Structurally, optimal lattices were characterized by low relative density (i.e., lightweight), high mechanical strength, and effective integration between the core and face sheets to facilitate efficient load transfer. Additionally, lattice designs that aligned with or extended findings from prior published studies were considered more valuable for comparative analysis.

The specimens produced from the two build configurations are displayed in Figure 5.8. Overall, manufacturability was not hindered by overhang resolution or surface defects; rather, the primary differentiator within this category was the ease of powder evacuation during post-processing. Specimens with higher core densities approaching 20% presented significantly greater challenges in powder removal compared to those with densities near 10%. This trend was anticipated to intensify in full-scale sandwich beam specimens containing a greater number of unit cells, a prediction that was substantiated by the beam sample. Among the TPMS lattice types, the SplitP configuration exhibited the highest difficulty in powder removal, indicating potential limitations for its use in large-scale applications.

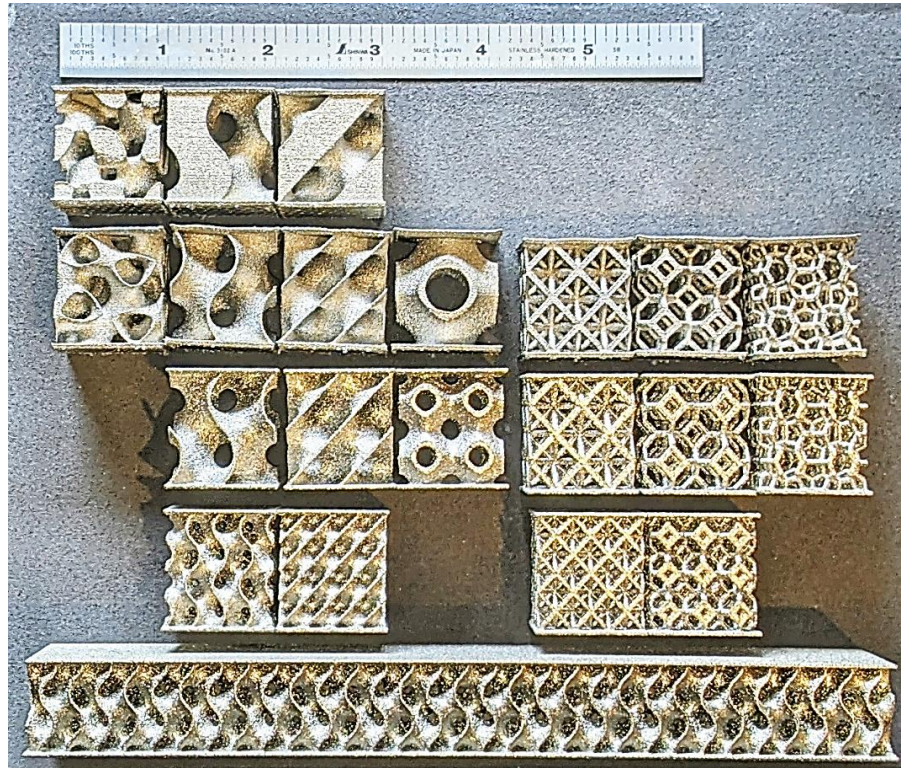


Figure 5.8. Sandwich lattice manufacturing quality specimens from builds 4 and 5.

The lattice configurations selected for evaluation were chosen primarily for their capacity to exemplify geometries uniquely enabled by additive manufacturing. Among these, the TPMS structures exhibited markedly greater geometric complexity and organic morphology compared to conventional strut-based lattices, owing to their continuous and smoothly varying surface curvatures. This inherent structural continuity not only enhances their aesthetic and architectural distinctiveness, but also contributes to superior multifunctional potential. Specifically, the subdivided internal volumes of TPMS lattices offer opportunities for integrated functionalities, such as the simultaneous transport of multiple fluid streams within a single structure. Those capabilities that are not feasible with strut-based configurations, which typically support only single-channel flow paths.

From a structural standpoint, low-density and lightweight lattice geometries are generally preferred for aerospace applications, where minimizing mass without compromising mechanical strength is critical. In this context, strut-based architectures demonstrated superior performance relative to TPMS configurations. For instance, a Kelvin cell lattice comprising 2^3 unit cells exhibited an equivalent relative density to a single unit cell Diamond TPMS structure. The continuous surface curvature and inherently thin wall profiles of TPMS lattices impose limitations on achieving ultra-low relative densities, potentially rendering them less desirable for weight-sensitive applications.

Mechanical strength across all configurations was found to correlate directly with relative density, as no significant defects or stress concentrations were observed. Consequently, TPMS skeletal structures outperformed others in terms of strength-to-density ratio; however, their comparatively higher mass conflicted with the research objectives focused on lightweight design, leading to their exclusion from further experimental evaluation. Regarding the integration of face sheets, all core configurations demonstrated adequate compatibility except for the Weaire–Phelan structure. This particular lattice exhibited numerous small, nearly orthogonal contact interfaces with the face sheets, which are suboptimal for efficient load transfer. Such geometric constraints hindered its structural integration and limited its applicability in sandwich panel designs.

In terms of comparative relevance to existing literature, five lattice configurations emerged as the most frequently studied and referenced: the strut-based Octet and Kelvin structures, along with the TPMS-sheet variants Gyroid, Diamond, and Primitive. These configurations not only aligned with established research but also performed favorably across the qualitative assessment criteria, particularly at lower relative densities. Post-processing considerations, specifically powder removal, played a significant role in the final evaluation. To ensure manufacturability and

minimize post-processing complexity, target relative densities for sandwich beam specimens were constrained to approximately 10%. Within this range, strut-based lattices demonstrated markedly superior powder evacuation characteristics compared to TPMS-sheet structures.

A comprehensive summary of the evaluated lattice sandwich designs is presented in Table 5.7. Based on their favorable performance across manufacturability, additive manufacturing suitability, structural integrity, and literature relevance, the Kelvin, Octet, Diamond, Gyroid, and Primitive configurations were selected for continued investigation in subsequent experimental phases.

Table 5.7. Summary of lattice manufacturing quality assessment

Type	Unit Cells	ρ	over-hang	defects	PRS	make	organic	multi-function	low-density	strength	integrate	Lit.
Kelvin	2	0.09	0	-	+	+	-	-	+	0	0	+
Kelvin	3	0.18	0	0	-	+	-	-	-	+	0	+
Octet	2	0.17	0	0	+	+	-	-	0	+	0	+
Octet	3	0.19	0	0	0	+	-	-	-	+	0	+
Weaire-Phelan	2	0.14	0	0	-	+	-	-	0	0	-	-
Diamond	1	0.09	+	0	+	+	0	+	+	0	0	+
Diamond	2	0.19	+	0	-	+	0	+	-	+	0	+
Gyroid	1	0.08	+	0	+	+	+	+	+	0	0	+
Gyroid	2	0.16	+	0	-	+	+	+	-	+	0	+
Primitive	1	0.14	+	0	0	+	+	+	0	0	0	+
Primitive	2	0.2	+	0	-	+	+	+	-	+	0	+
SplitP	1	0.11	+	0	-	+	+	0	+	0	0	-
Diamond	1	0.5	+	0	+	+	0	-	-	+	+	-
Gyroid	1	0.5	+	0	+	+	+	-	-	+	+	-
Lidinoïd	1	0.38	+	0	-	+	+	-	-	+	+	-

Another component of the manufacturing quality assessment was the orientation of the face sheets. In version 1 of the lattice build configurations, the face sheets were built parallel to the build plane. That orientation required support features across the lower surface resulting in poor quality, while the upper face sheet deformed around the periphery from insufficiently supported

corners and edges. The layout for version 2 rotated the face sheets 90 degrees to be parallel with the build direction, in which case support features were placed along the lower edges, and after removal were not noticeable. With both face sheets oriented vertically, they were identical in quality, had a uniform surface condition, and no observable defects. As for the lattice cores, there were no noticeable orientation effects due their symmetry. From these results, it was concluded that the optimal build orientation for the beam specimens would have the face sheets parallel to the build direction, i.e. perpendicular to the build plane.

5.2.5 *Initial Material Characterization*

Material behavior was characterized through tensile and compressive testing of specimens fabricated using non-standard, waste L-PBF Ti-6Al-4V powder. Specifically, nine ASTM E466-compliant tensile specimens with a nominal diameter of 9 mm, sourced from Build 5, were tested to failure under uniaxial monotonic loading in the as-built condition, with no post-processing. Additionally, nine cylindrical segments (8 mm in height) extracted from the grip sections of ASTM E466 9 mm diameter tensile specimens from Build 3 were subjected to compressive loading up to 250 kN. The resulting stress-strain data from both tensile and compressive tests are presented in Table 5.8 and Figure 5.9. In the tensile dataset, specimen nomenclature reflects spatial positioning within the build volume to account for potential variability. Designations include F (front), M (middle), B (back), L (left), and R (right), and additional information related to the material characterization can be found in Appendix D.

Table 5.8. As-built mechanical properties of Ti-6Al-4V produced by EB-PBF.

Measure	Units	Average	Std Dev	AMS 4911
Elastic modulus	GPa	104.74	3.96	
Elongation	%	18.00	1.53	
Yield tensile stress (0.2% offset)	MPa	913.4	24.2	868.7
Ultimate tensile stress	MPa	993.5	9.3	923.9
Tensile stress at failure	MPa	872.8	24.7	
Tensile strain (max load)	mm/mm	0.065	0.003	
Compressive modulus	MPa	8725.7	357	
Yield compressive stress	MPa	1065.7	25.5	
Compressive stress (max load)	MPa	2019.0	16.8	

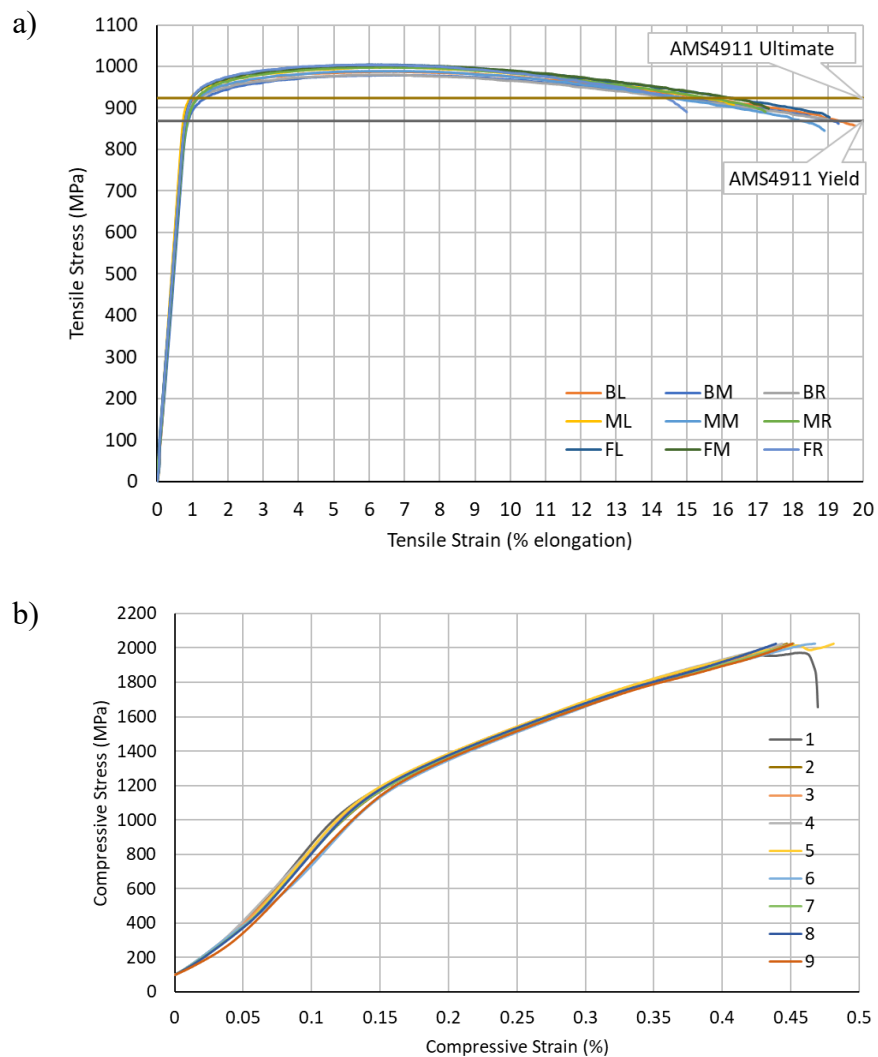


Figure 5.9. Stress-Strain plots from static a) tension, and b) compression, experiments with specimens produced from non-standard L-PBF Ti-6Al-4V powder.

The tensile test results revealed a consistent linear elastic response across all specimens, with an average 0.2% offset yield strength of 913.4 MPa. This value exceeds the nominal yield strength specified in AMS4911 for Ti-6Al-4V sheet stock (868.7 MPa) by approximately 5.1%, indicating enhanced mechanical performance relative to aerospace-grade standards. At an average peak load of 61.5 kN, the material exhibited an ultimate tensile strength of 993.5 MPa, representing a 7.5% increase over the AMS4911 benchmark. The calculated Young's modulus was 104.7 GPa, and the average elongation at failure was 18%, demonstrating excellent ductility. The narrow margin between yield and ultimate strength is symbolic of the alloy's characteristic near-perfect plastic deformation behavior, with only minor post-yield softening (~12%) observed at failure.

Compressive testing of the nine cylindrical specimens yielded an average 0.2% offset compressive yield strength of 1065.7 MPa, which is 9.8% higher than the typical value of 970 MPa reported for Ti-6Al-4V Grade 5. Notably, specimens 1 and 5 experienced fracture immediately prior to reaching the test apparatus's maximum load capacity of 250 kN. To account for initial surface irregularities, specifically residual support structures on the specimen bases, the stress-strain curves were normalized to zero strain at 100 MPa. This adjustment mitigated artificially elevated strain readings at low loads, ensuring accurate representation of bulk material behavior during platen engagement.

Figure 5.10 presents both optical and scanning electron microscopy (SEM) images of the fractured tensile specimen (MM) and compressive specimen (1). The tensile specimen exhibited classic ductile fracture characteristics, including pronounced plastic deformation and the formation of a "cup-and-cone" topography. Observable necking was present, and the fracture surface displayed a fibrous, rough texture with numerous dimples, indicative of microvoid coalescence, a hallmark of ductile failure mechanisms.

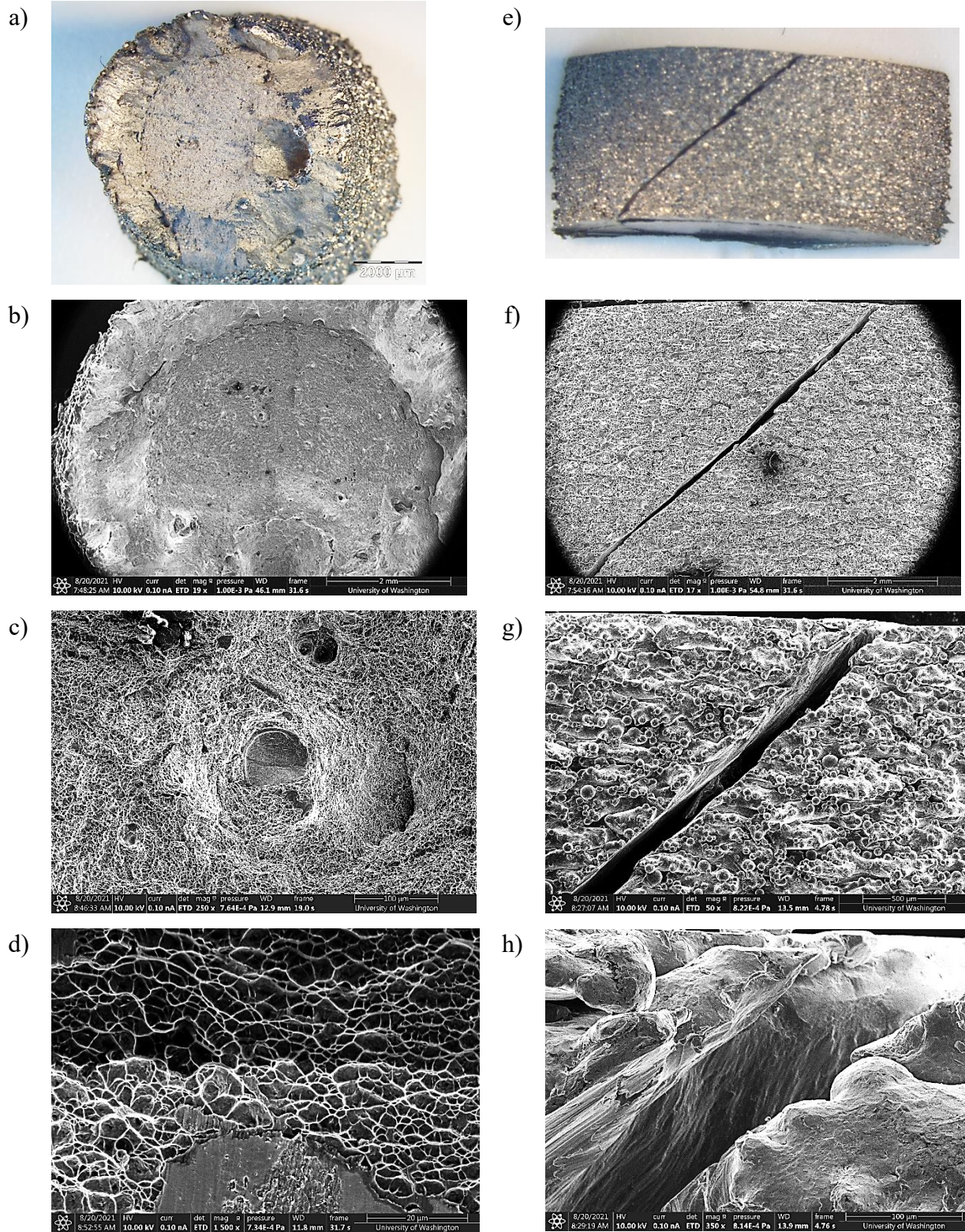


Figure 5.10. Optical and SEM images of the fracture surface from tension specimen MM a) through d), and compression specimen 1 e) through h).

In contrast, the compressive specimen demonstrated outward bulging at the periphery, with failure manifesting along a diagonal fracture plane oriented approximately 45 degrees to the loading axis. This fracture mode is consistent with shear-dominated failure, where shear stresses concentrated along the inclined plane induced material displacement. Consequently, the internal surface of the compressive fracture appears relatively smooth, in stark contrast to the rough, dimpled morphology observed in the tensile fracture surface.

5.3 SANDWICH STRUCTURE DESIGN AND BUILD

The initial phase of this study involved the comprehensive characterization of Ti-6Al-4V powder discarded from a L-PBF process, with its chemical composition verified to conform to established material standards. Subsequently, process parameters were identified to enable the effective integration of this non-conventional powder feedstock into the EB-PBF manufacturing workflow. The third phase entailed a targeted assessment of lattice core topologies, namely Kelvin, Octet, Diamond, Gyroid, and Primitive structures, to determine those offering optimal structural integrity and manufacturability within sandwich panel architectures. Then, mechanical property data were generated to support finite element modeling, enabling topology optimization and predictive performance evaluation. Accordingly, the ensuing section presents the outcomes of the next phase, which encompassed the design, fabrication, and analysis of lattice sandwich structures informed by the preceding investigations.

Initially, a parametric model framework was established using nTopology, the details of which are in Appendix B. This software program not only provided the ability to model the complex lattice core configurations, but to also mesh the models for FEA and topology optimization. The ability to mesh and simulate the native model geometry without file type conversions and simplifications significantly streamlined the DfAM workflow. Once a generic

sandwich beam model was set up for each core type, three different variations of each lattice core configuration were designed. One would have a uniform lattice thickness, and the other two would have a continuously variable thickness based either on a SIMP-based topology optimized density field or a von Mises stress field. For all cases, unit cell dimensions were held constant, and thickness parameters were calibrated to achieve an approximate relative density of 10% within the lattice cores.

5.3.1 *Lattice Core Modeling*

Design of the sandwich structures began with the generic layout of the beam. The three main components were the upper face sheet, lower face sheet, and the lattice core volume. Design of the sample beam, fabricated during the lattice manufacturing trial in Section 5.2.3, attempted to adhere closely to the description defined in ASTM C393. The length of the beam was limited by the Arcam A2X build area and set to 190 mm. That also set the span length to 140 mm which was specified to be 50 mm less than the total length. The cross section was to be rectangular with a depth equal to the thickness of the sandwich construction. In terms of the specimen width, it had to be greater than twice the thickness (50.8 mm), not less than three times the dimension of a core cell (i.e., 3 or more unit cells across), nor greater than half the span length.

From that description of width requirements, there were several dependencies based on cell size, beam width, and core depth. That meant the core depth was decreased from the 25.4 mm, used for the cube samples, to 19.05 mm, and that set the width to 38.1 mm. In order to have three unit cells across the width, the cell size was set to 12 mm x 12 mm x 12.7 mm (L x W x H). With a Gyroid cell lattice thickness of 1 mm, the relative density of that configuration was 16%. After fabrication of that sample, the post-process powder removal was found to be challenging and far too labor intensive to use those same parameters for the 4-point bend experimental beam

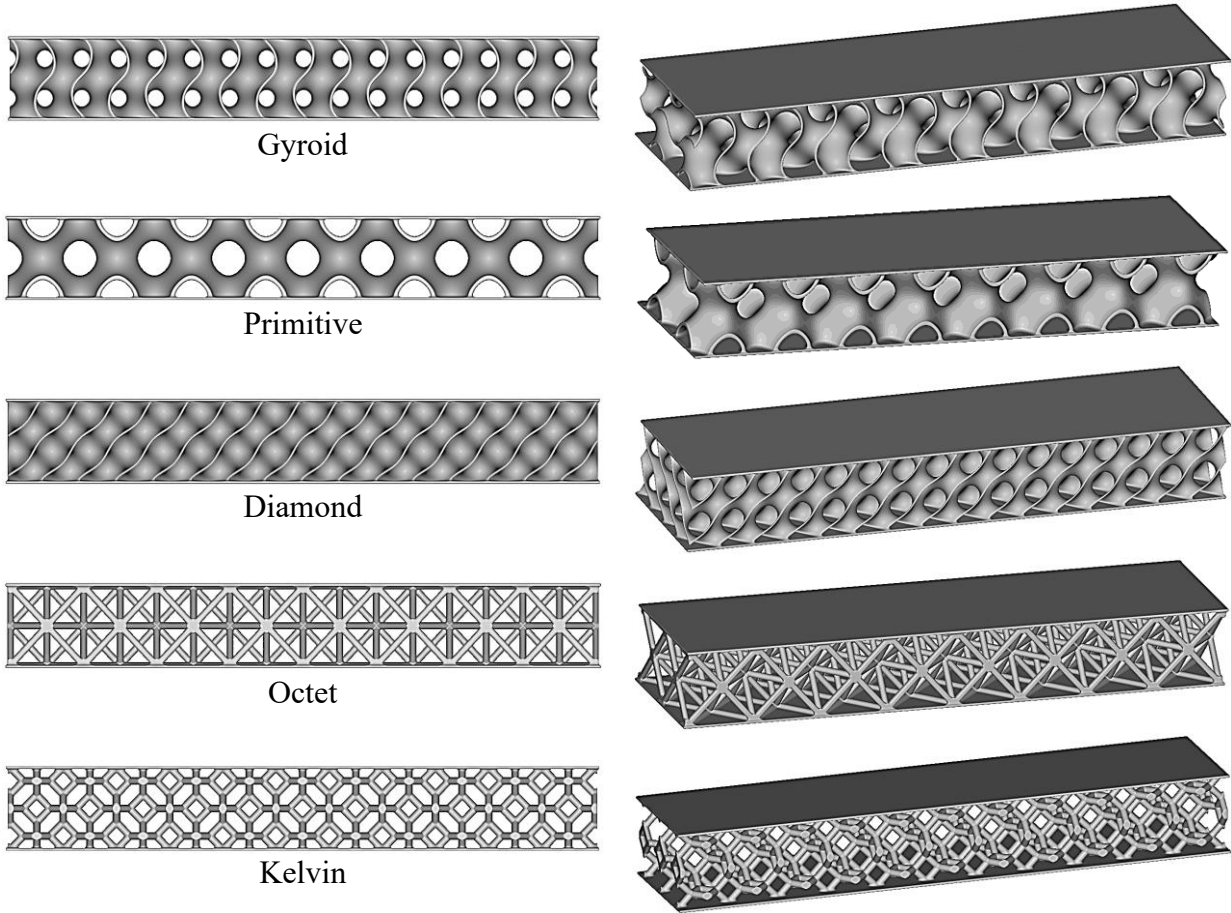
specimens. Therefore, it was necessary for the beam design to deviate from the ASTM C393 definition of a “standard configuration” to what the specification defined for a “non-standard configuration.”

The unit cell size was increased to 24 mm x 24 mm x 25.4 mm (L x W x H) to maintain larger passageways for powder removal via the PRS. Unit cell quantity across the width was decreased from three to two, which set the width of the beam to 48 mm, and drove a slight increase to the beam length up to 192 mm. It should be noted that for meshing reasons the Kelvin and Octet unit cell sizes had to be adjusted slightly to 17.2 mm x 18.6 mm x 17.6 mm, and 23.6 mm x 22.4 mm x 25.4 mm, respectively. This was necessary to avoid thin, partially formed struts at the boundaries of the core volume, which were driving minimum mesh element sizes too small to be adequately resolved. Considering the highly unique core structures, these changes were deemed to be reasonable, and not significant enough for the experimental data to not be comparable to other metallic sandwich structure. Other design parameters such as face sheet thickness (0.8128 mm) and the face sheet-to-core blend radius (0.8 mm) were held constant.

These design values established the initial blocks of the overall beam modelling framework. From there, the core geometry was defined by preprogramed volume lattice fill types: Gyroid, Diamond, Kelvin, Octet, and Primitive. To establish the baseline uniform core sandwich beam configurations, with the unit cell size fixed to ensure a quality structure, the parameter for lattice thickness was tailored for each core geometry to drive the calculated relative density to ~10%. Differences in surface areas of the various configurations meant no two had the same design thickness. For the uniform core structures, the strut types Kelvin and Octet, had lattice thicknesses of 2.4 and 2.2 mm, respectively, whereas for the TPMS types, Gyroid, Diamond, and Primitive,

lattice thicknesses were set to 1.3 mm, 1.1 mm, and 0.75 mm, respectively. The resulting lattice core sandwich beam models are shown in Table 5.9.

Table 5.9. Side and isometric views of the five sandwich structure beams designed with uniform thickness TPMS-type and Strut-type lattice cores.



Following the general design layout of the specimens, the geometries were meshed in order to run FEA and topology optimization. In nTopology, the method constructed for meshing an implicit body was to: 1) turn the model into a voxel grid of tiny cubic elements (e.g. 0.3 mm), 2) convert the voxel grid into a grid mesh, 3) convert the grid mesh to a triangulated surface mesh, and finally 4) create a volume tetrahedron mesh based on the surface mesh nodes. Mesh parameters at every step varied slightly for each configuration, and were set in order to sufficiently

capture the model details with a desire for more than one element through the thicknesses. However, there was a computing resource limitation that did not allow models to exceed approximately 1.2 million elements. An example of a voxel grid and triangle surface mesh for a Gyroid lattice is shown in Figure 5.11.

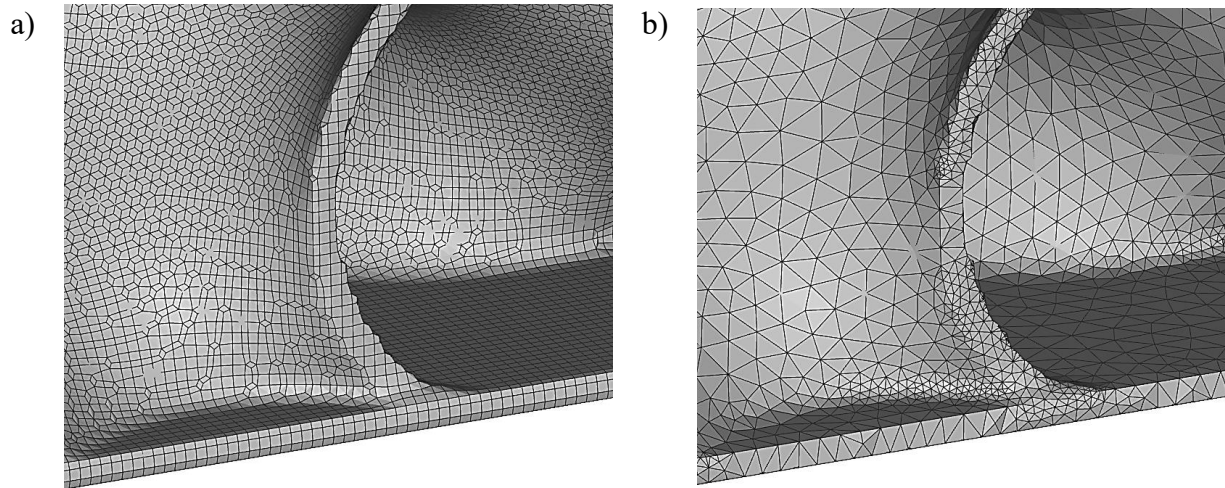


Figure 5.11. Example of converting an implicit body resembling a Gyroid lattice to a) a voxel grid mesh, and then to a b) triangulated surface mesh.

In order to setup the 4-point bend simulations for predicting the performance of the structures, the material properties, loads, and boundary conditions, had to be defined. For a simple static analysis, only the Young's modulus (104.7 GPa), Poisson's ratio (0.342), and density (4.43 g/cm³), were required to define the elastic material properties. Loading configuration was defined by ASTM C393 for 4-point third-span loading as shown in Figure 5.12, which placed the load across the beam width at 1/3 and 2/3 the span length (142 mm). The supports were modeled across the width as cylinders intersecting the beam mesh surface, and any encompassed nodes were restrained with fixed displacements, while rotation in the width (Y) direction was allowed. Typically, it is

desired to minimize restraints; however, the model would not run a static analysis with fewer constraints due to the detection of rigid-body motion.

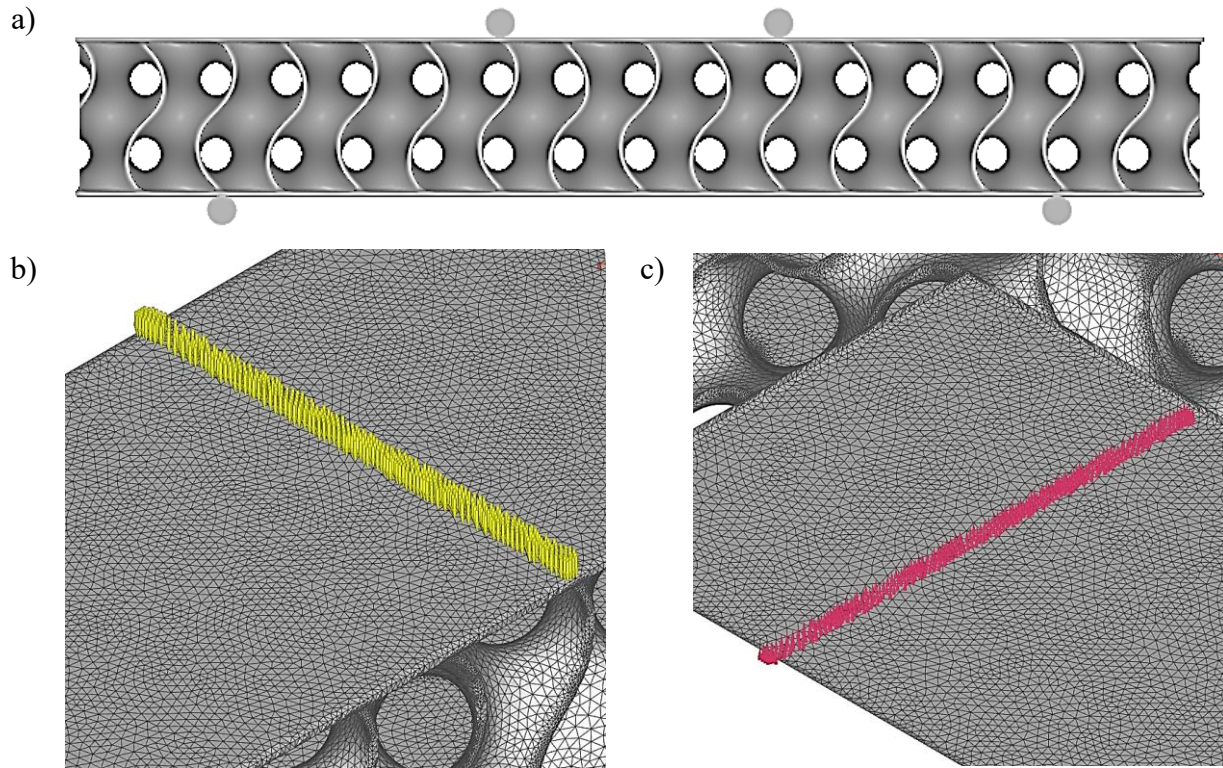


Figure 5.12. a) Beam model configuration for 4-point bend loading simulation, b) example of how load inputs are applied at FEM nodes across the beam width, c) example of how boundary constraints are applied at FEM nodes across the width.

5.3.2 *Topology Optimization Grading*

The modeling framework developed for uniform lattice core configurations provided a sufficient basis for initial design and simulation tasks. This baseline model subsequently served as the foundational geometry for the development of graded-thickness lattice cores. Two distinct sets of five sandwich beam specimens were generated: one set utilized a density gradient field derived from SIMP-based topology optimization, while the other employed a von Mises stress field obtained through finite element static analysis. Both gradient fields were informed by a 4-point bend load case simulation. However, a critical unknown remained regarding the appropriate

definition of the topology optimization design space, specifically whether it should be represented as a homogeneous solid block or a structurally equivalent thickened lattice domain.

In conventional applications of the SIMP method, the design space is typically defined as an enlarged solid domain. This affords the optimization algorithm maximum geometric freedom in identifying optimal material distributions under prescribed loading conditions, with minimal user-imposed constraints. For the present study involving sandwich beam structures, the design space was delineated as the full rectangular volume situated between the upper and lower face sheets, corresponding to the region occupied by the lattice core. A volume fraction constraint of 10% or less was imposed to ensure material efficiency, and the objective function was formulated to minimize structural compliance under a simulated 4-point bend loading scenario.

After 34 iterations the optimization converged, meaning all constraints were satisfied, in addition to the convergence criteria of 0.003 minimum objective change, and 0.03 minimum density change. The meshed design space and resulting optimized shape are displayed in Figure 5.13a and Figure 5.13b. The topology optimization algorithm removed material from where it was not needed to satisfy the simple load condition, and maintained material between the load input and support locations. When the result was viewed as a density field in Figure 5.13c, the arch structure is clearly visible with fully dense material in the middle and pseudo-density material growing outwards. When that density field was mapped to an implicit TPMS Diamond lattice thickness gradient, that result is shown in Figure 5.13d.

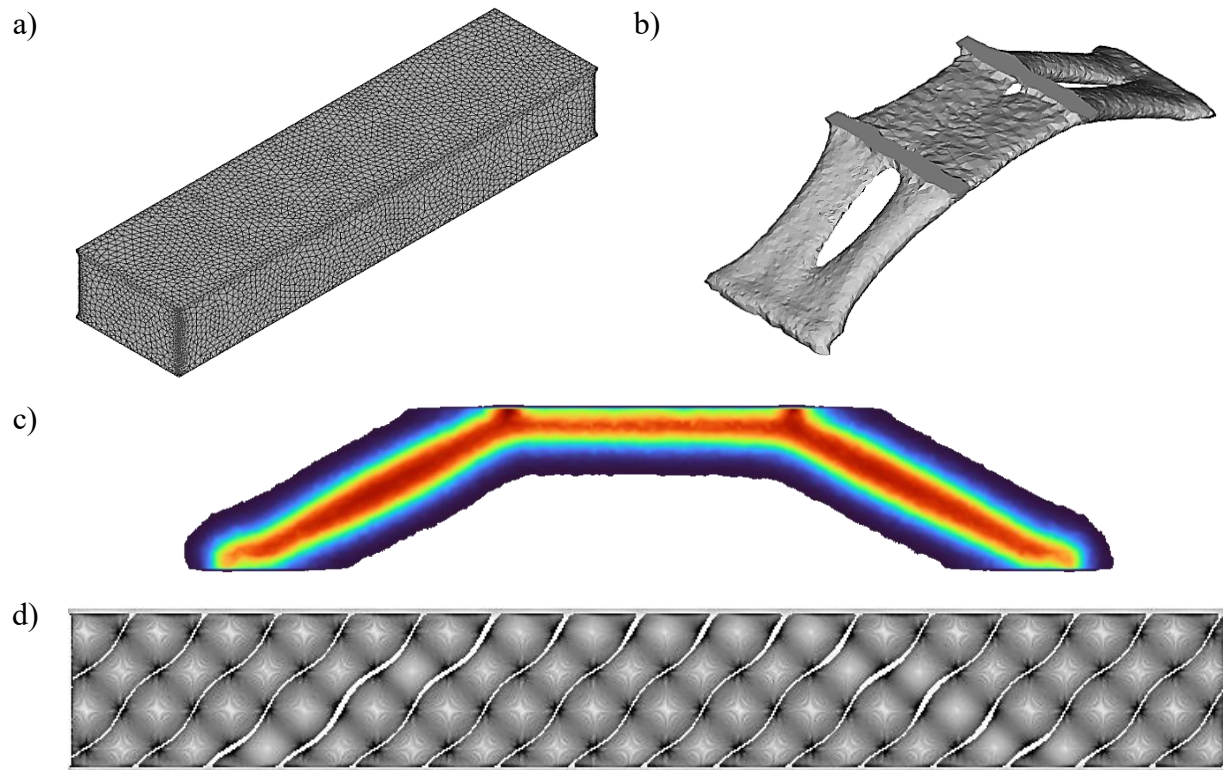


Figure 5.13. a) Design space meshed with linear tetrahedron elements, b) topology optimized resulting shape, c) cross-section of resulting density field, d) cross section of density field applied to thickness of Diamond lattice beam.

The shape itself was perhaps an obvious result with material placed on straight paths between the load input points and the supports; however, rather than being uniform across the width, it was divided into two arches joined at the middle and the bases. Instead of a single thinner uniform cross section in the arch legs, the division allowed for increased thickness to provide a larger second area moment of inertia, which reduced the compliance given the mass constraint. While the optimization did not produce a lattice configuration, a resulting density scalar field was applied as a modifier to a lattice thickness variable. In this way the lattice thickness was continuously graded on a linear scale between minimum and maximum values for thickness and density. For example, the density varied from 0 to 1 (void to solid), so wherever the density field was 0 the

lattice thickness was defined to be minimal at 0.6 mm thick, and vice versa where the density field was 1 the lattice thickness was set to a maximum value of 1.5 mm. Pseudo-density regions with values between 0 and 1, the lattice thickness was graded on a linear scale from 0.6 mm to 1.5 mm.

Nevertheless, simply transferring a density or stress field from a solid volume to a porous lattice geometry did not make sense. This approach was flawed in that there are already voids in locations where the optimization says the material should be fully dense. The load path from the inputs cannot follow straight lines to the support boundaries and must take twisted paths through the complex structure. If the performance of the core was to be enhanced by optimizing thicknesses along the key load paths while maintaining the overall lattice geometry, then the field modifying the thickness must be derived from a simulation of a design space resembling the desired lattice core structure. Therefore, rather than applying a single density or stress field to the five different lattice cores to be tested, a separate design space was required for each of the five lattice configurations.

Design spaces for the graded core structures, were derived from models of the uniform geometries, but with a significantly increased thickness. For the SIMP topology optimization runs, the design space thicknesses were set to 4 mm with a 20% volume fraction constraint applied. The intent of the lattice core topology optimization was not the resulting shape of the core where material would be removed, but rather the density field showing where material thickness should be increased or decreased to minimize the overall compliance. To obtain a stress field thickness modifier, a linear static 4-point bend analysis was run on a uniform core sandwich structure with a 30kN load applied. For TMPS-type cores the web thickness was set to 2 mm, and for strut-type cores the strut diameters were set to 3 mm. The resulting von Mises stress maps from each different core geometry were applied to the thicknesses with tailored stress values for minimum

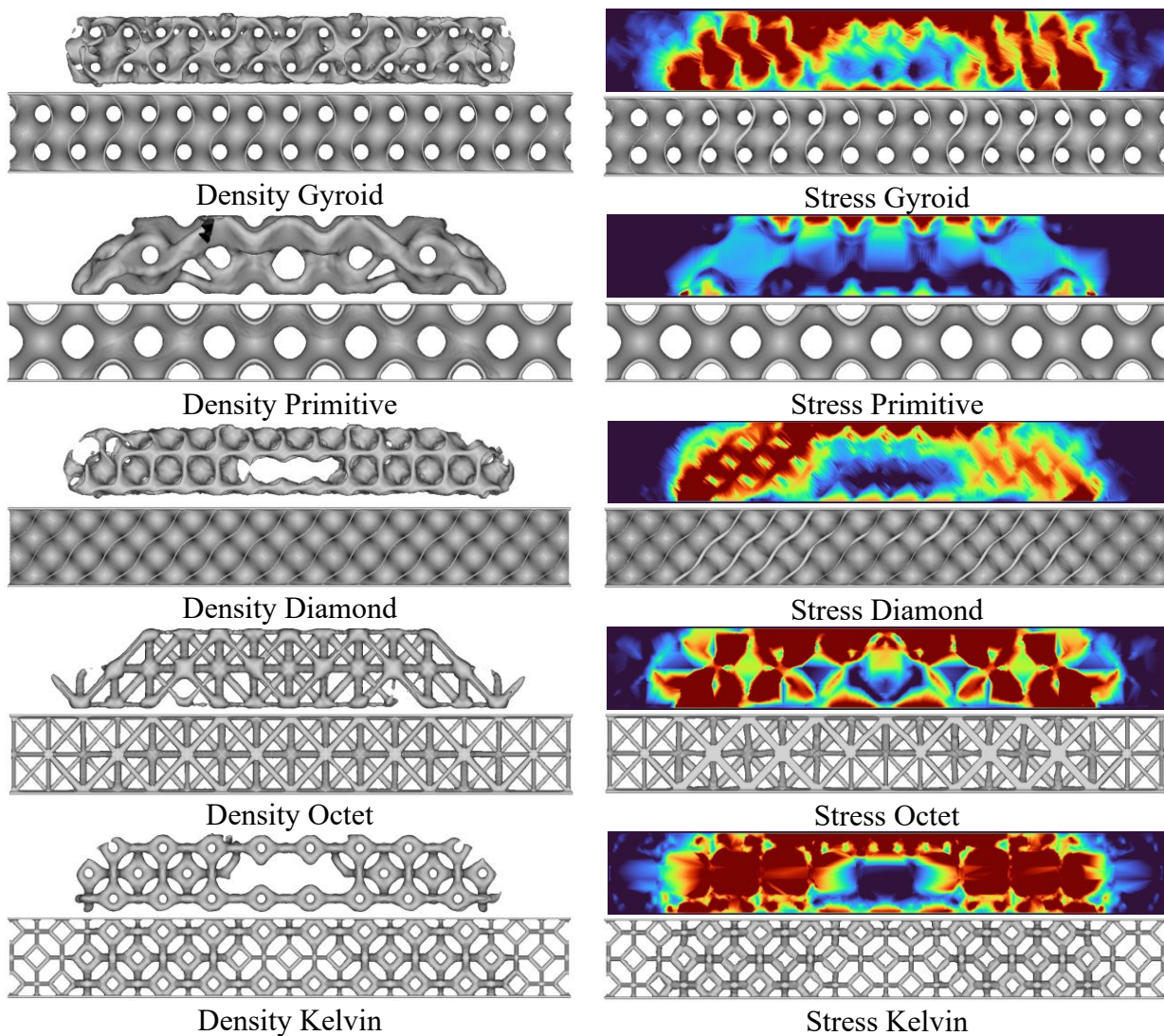
and maximum thickness values. Table 5.10 shows the design parameters and resulting relative densities of the fifteen different sandwich structures.

Table 5.10. Design thickness parameters for all lattice core configurations

Grading Method	Lattice Type	Lattice Name	Lattice	Sim Lattice				Graded Lattice			Lattice Rel. Den.
			Thickness mm	Thickness mm	Min Thick mm	Min Value	% min	Max Thick mm	Max Value	% max	
Uniform	TPMS	Gyroid	1.3	-	-	-	-	-	-	-	0.103
Uniform	TPMS	Primitive	0.75	-	-	-	-	-	-	-	0.106
Uniform	TPMS	Diamond	1.1	-	-	-	-	-	-	-	0.105
Uniform	Beam	Octet	2.2	-	-	-	-	-	-	-	0.105
Uniform	Beam	Kelvin	2.4	-	-	-	-	-	-	-	0.108
Density	TPMS	Gyroid	variable	4	0.6	0.05	-	2.0	0.45	-	0.101
Density	TPMS	Primitive	variable	3	0.6	0.40	-	2.0	1.00	-	0.104
Density	TPMS	Diamond	variable	4	0.6	0.05	-	2.0	0.60	-	0.103
Density	Beam	Octet	variable	4	1.5	0.05	-	3.5	0.55	-	0.103
Density	Beam	Kelvin	variable	4	1.5	0.05	-	3.5	0.45	-	0.109
Stress	TPMS	Gyroid	variable	2	0.6	2.00E+07	13.8	2	2.00E+08	26.7	0.107
Stress	TPMS	Primitive	variable	2	0.6	4.00E+07	35.9	2	4.00E+08	2.7	0.109
Stress	TPMS	Diamond	variable	2	0.6	3.00E+07	22.4	2	2.50E+08	13.3	0.109
Stress	Beam	Octet	variable	3	1.5	2.41E+07	13.6	3	2.00E+08	27.3	0.111
Stress	Beam	Kelvin	variable	3	1.5	3.00E+07	14.3	3	3.00E+08	28.2	0.104

Distinct differences between the stress-driven and density-driven lattice core designs are evident in the side-view profiles presented in Table 5.11. In the stress-graded configurations, elevated stress concentrations were observed near the boundaries of the lattice core, where limited surrounding support material resulted in localized reinforcement. Consequently, these regions were assigned maximum thickness values along primary load-bearing paths to accommodate the increased mechanical demand. In contrast, the SIMP-based topology optimization approach identified the peripheral regions of the core as having minimal influence on overall structural compliance. As a result, material distribution in these areas was minimized, leading to reduced thickness at the edges. This divergence in grading methodology is most clearly illustrated in the TPMS Gyroid and Diamond configurations, which exhibit pronounced differences in edge reinforcement between the two design strategies.

Table 5.11. Side views of the five sandwich structure beams designed with density and stress graded thickness TPMS-type and Strut-type lattice cores.



With the completion of these comparative design results, the development phase for the fifteen distinct sandwich structure configurations intended for fabrication via EB-PBF was finalized.

5.3.3 FEA Predictions

A 4-point bend, second-order static analysis was performed on each of the 15 configurations within the nTopology software platform. While this type of analysis does not predict non-linear plastic

behavior or failure, it does have the ability to predict trends. It also provides a comparison of the different lattice structures. In Figure 5.14, the von Mises stress plot comparison illustrates how each structure distributes the stress from the load points to the supports. The scale was set to the yield criteria of Ti-6Al-4V, so that everywhere at or beyond yield is red for the 30 kN applied load.

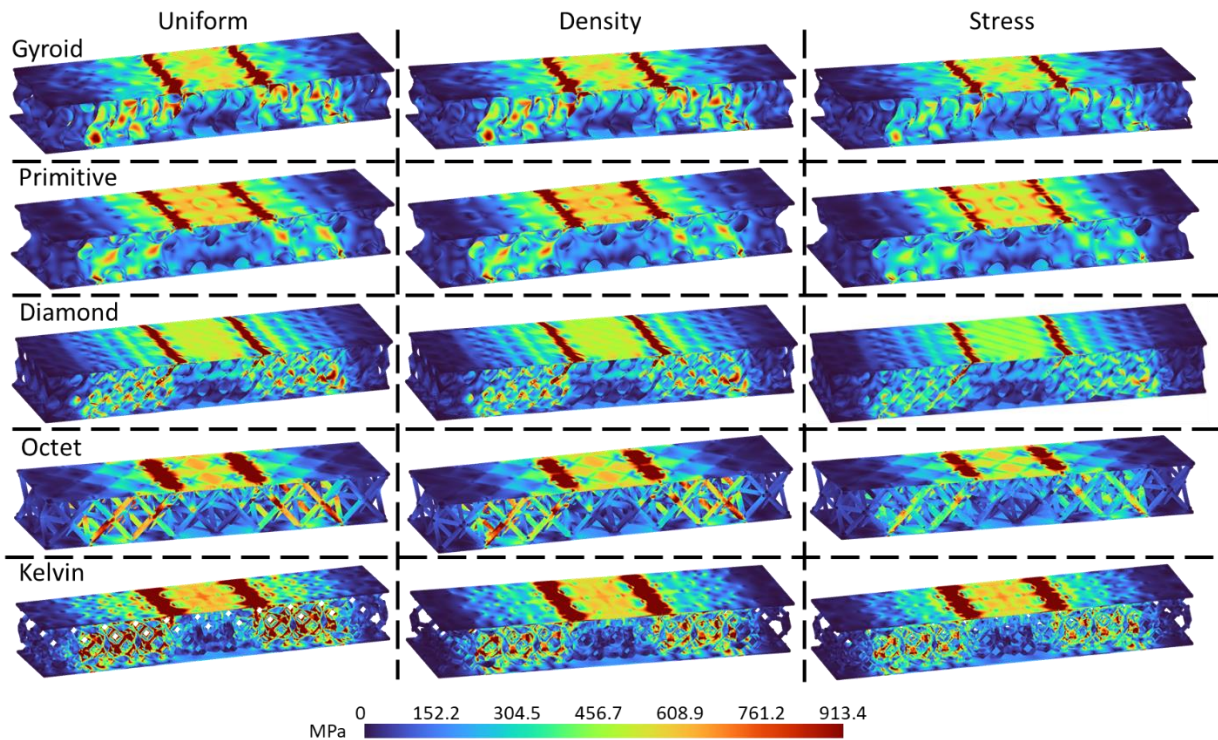


Figure 5.14. Finite element linear-static von Mises stress analysis under a 30 kN input load (scale set to YTS 913 MPa).

Excluding regions adjacent to the boundary conditions, where artificially elevated stress values may arise due to constraint effects, the graded lattice structures exhibit reduced overall stress magnitudes and more homogeneous stress distributions. These observations underscore the relative mechanical performance of the grading strategies: uniform grading demonstrates the lowest structural efficiency, followed by density-based grading, while stress-based grading yields

superior load-bearing capability. Among the TPMS configurations, the Diamond lattice exhibits fewer localized stress concentrations compared to the Gyroid and Primitive variants. Similarly, within the strut-based lattice category, the Octet structure presents lower stress levels and diminished stress localization relative to the Kelvin lattice. The stress contour plots further emphasize the effectiveness of TPMS architectures, particularly their curved web geometries, in facilitating more uniform bending load distribution across the structure.

Figure 5.15 presents a quantitative assessment of the predicted mechanical performance of the lattice sandwich structures. Given that linear static finite element analysis does not capture element failure or nonlinear material behavior, an alternative metric for evaluating load-bearing capacity involves monitoring the proportion of nodes exceeding the material yield criterion under increasing applied loads. Simulated loading conditions ranged from 10 kN to 60 kN across all fifteen lattice configurations, with results categorized according to the respective grading methodologies.

Independent of the applied grading strategy, the predictive finite element analysis identifies the Kelvin lattice topology as exhibiting the lowest structural performance, characterized by early onset of yielding and limited load-bearing capacity. For the remaining four topologies, a notable inflection point in the yield progression curves is observed between 30–40 kN under uniform grading, corresponding to approximately 2–3% of nodes exceeding the yield threshold. Within this load range, the proportion of yielding nodes approximately doubles and continues to accelerate with increasing load magnitude.

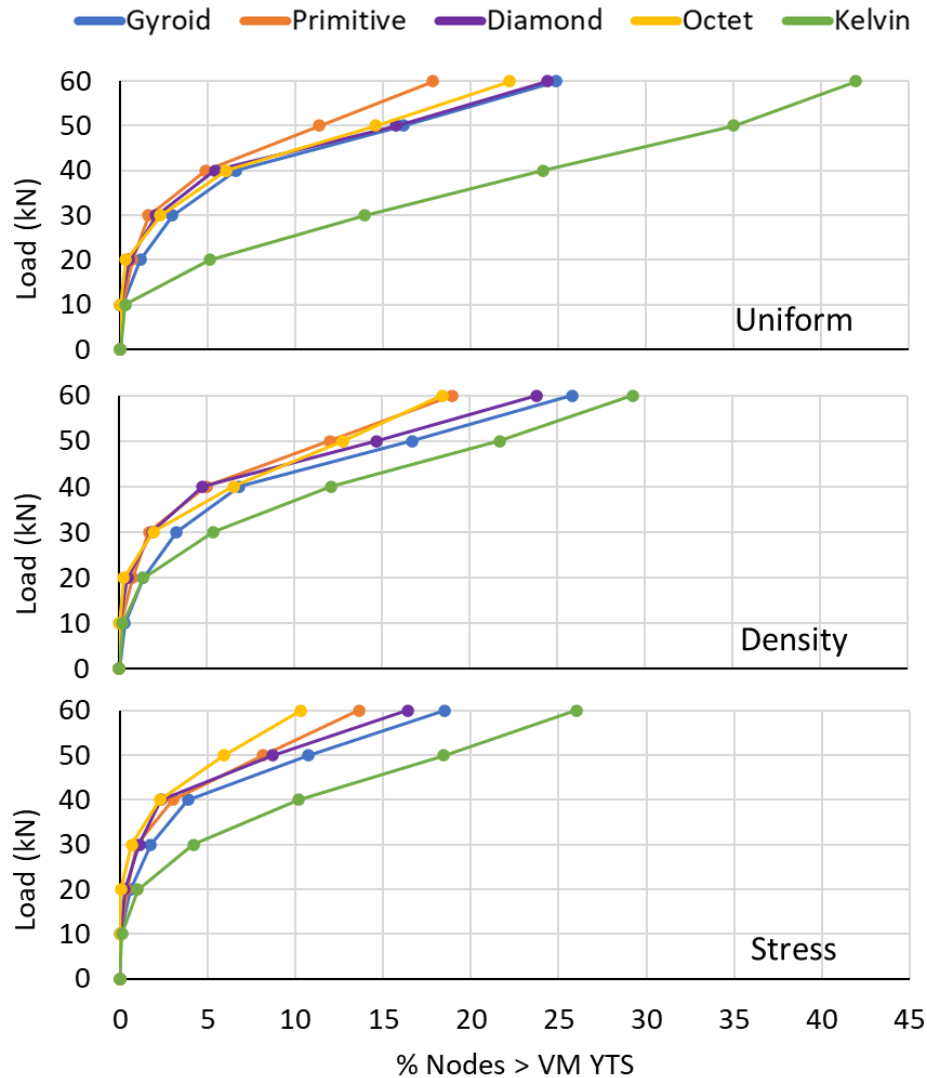


Figure 5.15. Plots showing the percentage of nodes in the finite element analysis that are exceeding von Mises stress yield criteria as a function of increasing load for the five different sandwich lattice cores and three grading methods (supporting data in Appendix B).

A similar trend is evident in the density-graded configurations, though the stress-graded designs demonstrate a delayed onset of rapid yielding, with the inflection point occurring between 40–50 kN. Under uniform grading, the Primitive lattice exhibits the greatest resistance to early yielding. In contrast, density grading enhances the performance of the Octet lattice, aligning its behavior with that of the Primitive. Under stress grading, the Octet surpasses the Primitive in yield

resistance, indicating a favorable response to stress-adaptive material distribution. Comparative analysis between the Gyroid and Diamond TPMS structures further reveals that the Diamond topology more effectively mitigates stress concentrations induced by bending, providing superior stress distribution characteristics.

5.3.4 *Moment of Inertia Analysis*

In classical beam theory, Euler and Bernoulli proposed a formula that described the deflection of a beam with a long aspect ratio, simply supported at both ends, under load. In a 4-point bend scenario, the deflection δ at various points along the beam are express by Equations (5.1)-(5.4),

$$\delta = \frac{Fx}{6EI}(3aL - 3a^2 - x^2), \quad 0 \leq x \leq a \quad (5.1)$$

$$\delta = \frac{Fa}{6EI}(3Lx - 3x^2 - a^2), \quad a \leq x \leq L - a \quad (5.2)$$

$$\delta_{max} = \frac{Fa}{24EI}(3L^2 - 4a^2), \quad x = \frac{L}{2} \quad (5.3)$$

$$\delta_a = \frac{Fa^2}{6EI}(3L - 4a), \quad x = a \quad (5.4)$$

where x is a distance along the supported length L , with two point loads F applied at a distance of a from the two supports, with the inverse of the elastic modulus E and the second moment of area I , also known as the moment of inertia. Classical beam theory demonstrates how important the second area moment of inertia term is, in which the EI term describes the flexural stiffness, or bending resistance of a structure. Since the Ti-6Al-4V bulk material is assumed to be isotropic, the elastic modulus E is constant; however, I in this case is continuously variable in all directions, not unlike bone structures [99].

In order to analyze the graded lattice designs, an open-source image computing platform called 3D Slicer (<https://www.slicer.org/>) was employed [199]. Slicer is a useful tool that can

perform many different analyses of STL models, like those used for 3D printing. Being open source, functions that are not built in can be developed and added as extensions by users. One such extension that was utilized is called SegmentGeometry, which is a Python-based extension created for analyzing CT scan data of bone structures, and it will plot the second moment of area along a particular axis. It accomplishes this by slicing a STL model into user specified increments, takes the solid cross sections as images, finds the second moment of area of each rectangular pixel, which is then summed to find the total for an arbitrary shape.

Each of the fifteen lattice designs were sliced into increments of approximately 0.25 mm along the length (I_{xx}), width (I_{yy}), and height (I_{zz}). The second moment of area was calculated according to Equation (5.5),

$$I_{na} = \sum_{i=1}^n I_{na'i} + A_i D_i^2 \quad (5.5)$$

where na is the neutral axis and na' is an arbitrary axis that crosses pixel i parallel to the neutral axis. As such, the second moment of area of the pixel about the arbitrary axis is represented by $I_{na'i}$, A_i is the pixel area, with D_i as the distance between the neutral axis and the arbitrary axis, and n being the number of pixels in the cross section. The output from this analysis is displayed in Figure 5.16.

Variations in the second moment of area (I_{xx}) along the beam length reveal distinct differences attributable to both the grading methodology and the underlying lattice core geometry. The uniformly graded lattice exhibits a relatively consistent pattern of alternating peaks and troughs, indicative of a repetitive structural response. In contrast, the density-graded and stress-graded configurations display more tailored stiffness profiles that reflect their respective optimization strategies under bending loads.

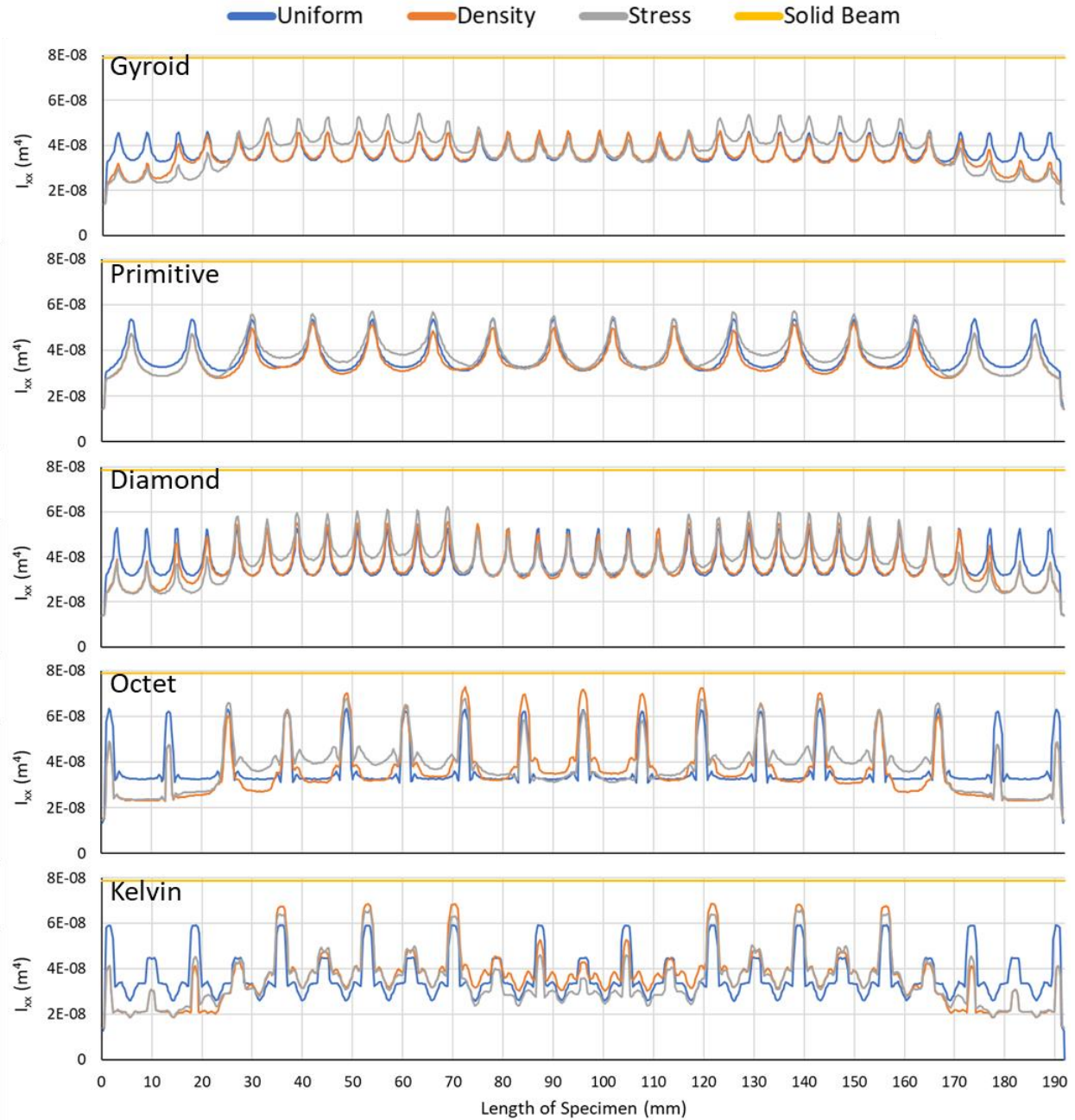


Figure 5.16. Second area moment of inertia I_{xx} analysis for lattice core sandwich beams.

At the beam extremities extending 25 mm beyond the support locations, stiffness was intentionally reduced due to the absence of significant structural loading. Notably, the density-graded structures exhibit a more gradual tapering of stiffness in these regions compared to the

sharper decline observed in the stress-graded counterparts. Between the support and load application points, both grading methods enhance bending resistance; however, the stress-graded designs demonstrate a more pronounced increase in stiffness, providing a more efficient material distribution in response to localized stress demands. Across the central span, where bending moments are minimal, both grading approaches reduce stiffness to conserve material without compromising structural integrity.

An intriguing observation arises in the TPMS Diamond stress-graded configuration, which displays an asymmetrical I_{xx} distribution along the beam length. This asymmetry is attributed to the inherent geometry of the unit cell, which includes angled load paths that align with the direction of force transmission toward the reaction supports. The side exhibiting a more direct load path experiences elevated stress concentrations, as confirmed by the stress field analysis, prompting the grading algorithm to locally increase stiffness. This behavior underscores the capacity of stress-based grading to exploit the anisotropic characteristics of TPMS geometries for enhanced structural performance. While only I_{xx} results were presented here, I_{yy} and I_{zz} results are located in Appendix B.

5.3.5 *Build Design*

Upon completion of the fifteen unique beam specimen designs, triangular surface meshes for each were exported from nTopology as STL files and loaded into Materialise Magics to arrange the build volume and add support features. The orientation of the beam specimens was established based on the manufacturing quality assessment, where the face sheets would be set perpendicular to the build plane. The overall configuration of the primary orientation horizontal build is shown in Figure 5.17 with the beams laid such that the long axis was parallel with the build plane. This was done as a risk mitigation in case of a build failure, such that any specimens that had been

completed prior to the failure would still be usable. Each column contained the same lattice core type (Primitive, Octet, Kelvin, Gyroid, Diamond), while the uniform graded started at the lower level, followed by the density graded, and the stress graded at the top.

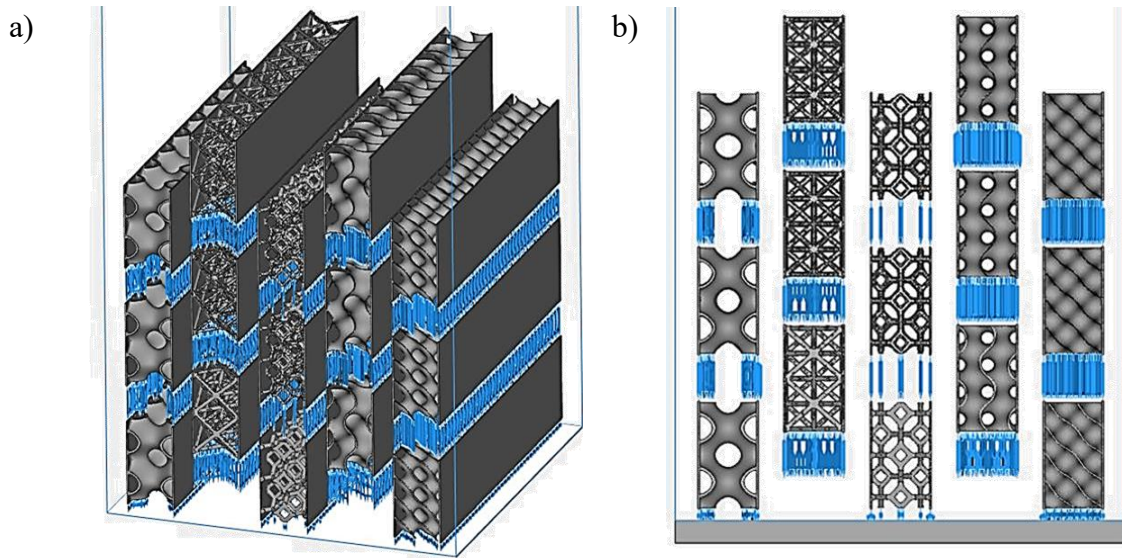


Figure 5.17. (a) Isometric and (b) side view of the horizontal lattice build layout within the Arcam A2X.

As a comparison, the alternative to the horizontal orientation, while keeping the face sheet perpendicular to the build plane, was to rotate the beam specimens 90 degrees to be built vertically. This was done to determine if there were any anisotropy in the sandwich structures related to build orientation. Figure 5.18 displays the vertical build layout, which resulted in a slightly shorter build at 197 mm (192 mm beam length plus 5 mm supports), and therefore less build time. The vertical layout also reduced the amount of support material, and thus time and labor required to remove them. However, in terms of powder removal there was no difference in difficulty compared to the horizontally built sandwich specimens.

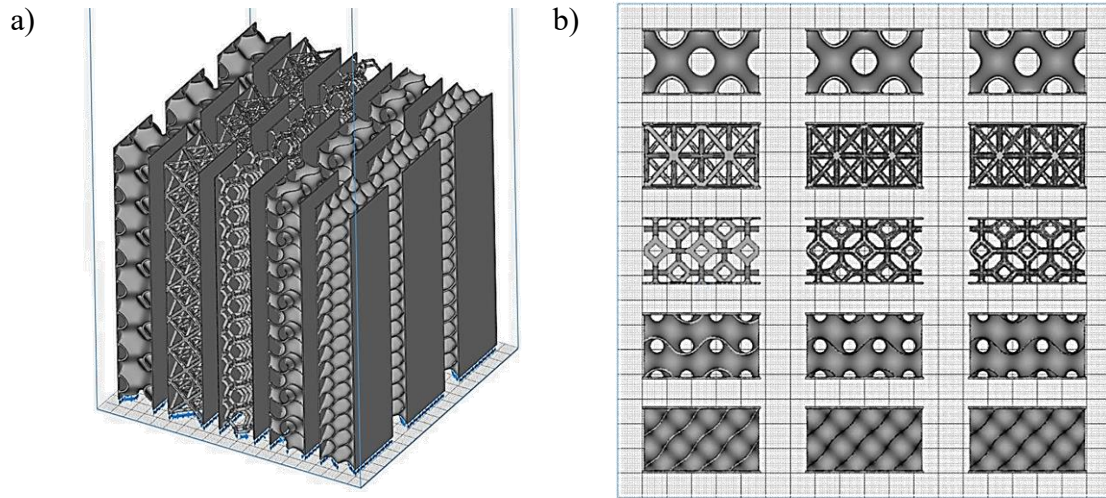


Figure 5.18. (a) Isometric and (b) top view (uniform, density, stress grading from left to right) of the vertical lattice build layout within the Arcam A2X.

In contrast to the specimens fabricated using waste L-PBF feedstock, a single representative of each sandwich beam configuration was manufactured in a horizontal orientation utilizing virgin Ti-6Al-4V powder specifically for EB-PBF, procured from Arcam. Due to the limited quantity of virgin powder available, it was not feasible to execute a single tall build encompassing all fifteen specimens. Consequently, the fabrication strategy was segmented into three discrete builds, as illustrated in Figure 5.19. Each build was dedicated to a specific grading scheme: the first comprised uniformly graded specimens, the second included density-graded specimens, and the third consisted of stress-graded specimens. All beam specimens were positioned with a 5 mm offset from the build plate and were affixed using support structures to ensure geometric and thermal stability during the build process.

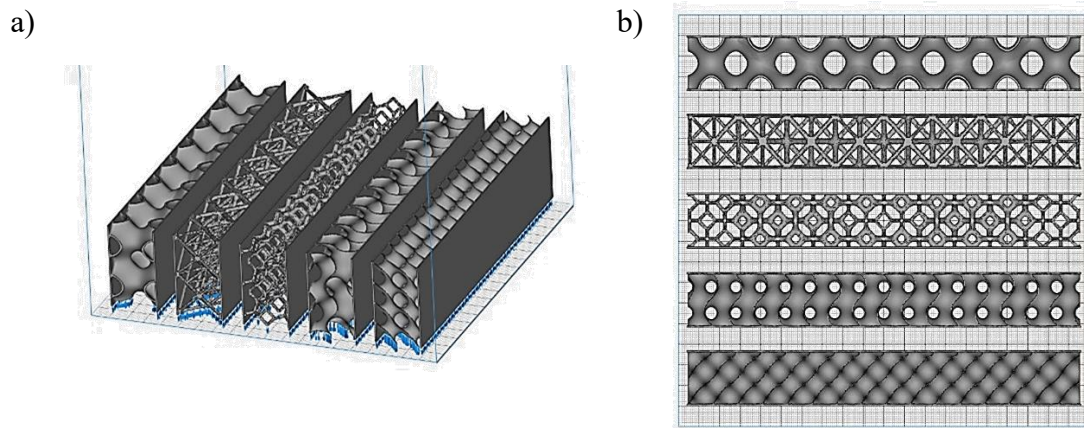


Figure 5.19. (a) Isometric and (b) top view of the lattice build layout for the beams fabricated with virgin EB-PBF Ti-6Al-4V powder.

To summarize, the build design strategy was carefully tailored to accommodate material constraints while reducing risk, and incorporating lessons learned from the initial manufacturing trials. By segmenting the fabrication into different builds, each corresponding to a specific orientation or material methodology, and ensuring consistent spatial orientation and support placement, the design phase established a robust foundation for comparative analysis. With the geometric configurations and build parameters finalized, the focus shifts to the subsequent stages of manufacturing and post-processing, where the physical realization of the specimens and powder removal will be addressed in detail.

5.3.6 *Manufacturing and Post-Processing*

Following the completion of the build layout in Magics, the finalized build models were exported to Arcam's proprietary slicing software for conversion into a machine-readable build file compatible with the Arcam A2X EB-PBF system. The machine setup replicated the configuration utilized in Build 5, including the use of the waste Ti-6Al-4V powder originally processed by L-PBF, which was loaded into the feed hoppers. The preheat protocol was consistent with prior

builds, initiating with a beam current of 10 mA. This approach effectively achieved the target build plate temperature of 730 °C without inducing powder overcharging or smoke events.

The build proceeded without interruption over a 66-hour continuous operation, culminating in a final height of 224 mm. Post-build operations, including component extraction, powder removal, and system clean-up, required approximately 18 labor hours, a factor of significance when evaluating the economic feasibility of such builds. Key post-processing activities, sintered powder cake extraction and PRS powder removal, are displayed in Figure 5.20. Upon completion of Build 6, the remaining filtered powder inventory was measured at 62.6 kg, indicating a net powder consumption of approximately 8 kg.

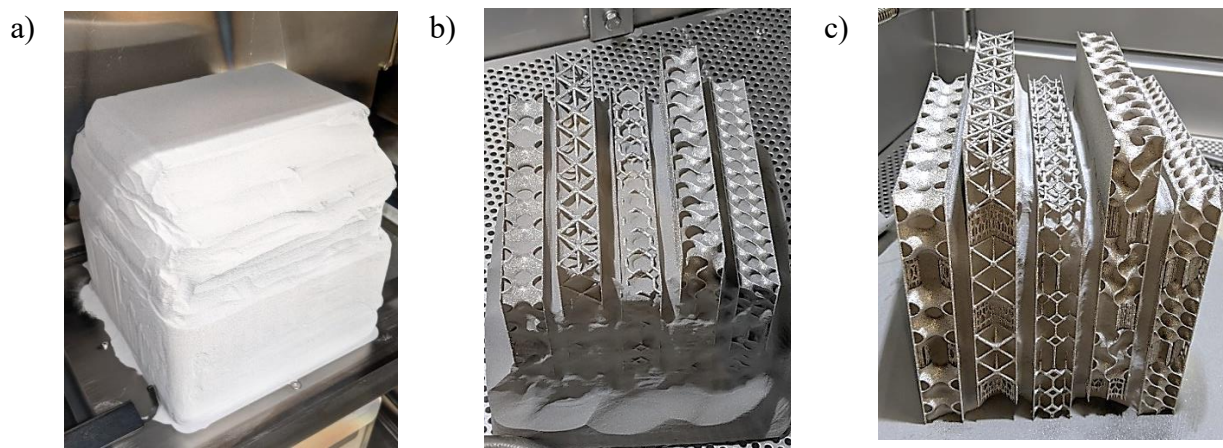


Figure 5.20. (a) Ti-6Al-4V sintered powder cake containing beam specimens, (b) Partial reveal of beam specimens embedded in sintered powder, (c) full reveal of all 15 beam specimens with sintered powder remaining between them.

In general, the manufactured sandwich beam specimens exhibited high fabrication quality, with no observable defects. However, post-processing challenges were notably more pronounced in the TPMS architectures, despite their relatively large unit cell dimensions and low target relative density. These difficulties underscore the limitations of the Arcam PRS, such that lattice core densities exceeding 10% are inadvisable when relying solely on this powder removal method.

Among all specimens, only a single manufacturing anomaly was identified, located on the upper-level stress-graded Primitive beam. As shown in Figure 5.21, the defect manifested as an unintended infill within a region that was geometrically defined to remain open, resulting in a localized melt on a single layer. This anomaly was isolated to one specific hole and did not appear in adjacent or similar features. No comparable defects were observed in other Primitive specimens or across different build levels. Despite thorough verification of the digital models and build files, the root cause of this irregularity remains undetermined. Interestingly, the same defect recurred in subsequent builds of the same configuration. Nevertheless, the affected specimens were retained for mechanical testing, with particular attention given to evaluating if the defect influenced the structural behavior.



Figure 5.21. Defect discovered on the stress-graded Schwarz Primitive beam specimen.

5.3.7 *Dimensional Analysis*

As part of the quality assurance protocol, dimensional verification of the fabricated specimens was conducted by comparing their geometries to the original CAD models. Given that the initial designs were generated using implicit field-based modeling techniques, such as those employed in

nTopology, direct dimensional interrogation is inherently challenging due to the absence of explicitly defined surface boundaries. To facilitate quantitative analysis, the implicit models were discretized and exported as explicit STL files, enabling more precise volumetric measurements. The nominal volumes derived from these STL representations, which served as the basis for fabrication, are presented in Table 5.12.

Table 5.12. Nominal model dimensions and measurements.

Lattice	Grading	Mass (g)	Volume (mL)	Density (g/cc)	Relative Density
Gyroid	Uniform	171.03	38.8	4.409	0.103
Diamond	Uniform	173.07	39.2	4.414	0.105
Kelvin	Uniform	174.93	38.4	4.556	0.108
Octet	Uniform	173.23	37.7	4.599	0.105
Primitive	Uniform	174.43	39.5	4.412	0.106
Gyroid	Density	168.89	38.3	4.415	0.101
Diamond	Density	170.32	38.6	4.409	0.103
Kelvin	Density	176.23	37.6	4.681	0.109
Octet	Density	171.29	37.6	4.554	0.103
Primitive	Density	173.17	39.1	4.425	0.104
Gyroid	Stress	175.07	39.7	4.413	0.107
Diamond	Stress	176.37	40.1	4.403	0.109
Kelvin	Stress	171.31	37.1	4.614	0.104
Octet	Stress	180.15	38.7	4.654	0.111
Primitive	Stress	178.08	40.3	4.417	0.109
Average Model		173.84	38.7	4.492	0.106

For the nominal mass values, a density for Ti-6Al-4V of 4.43 g/cc was input into the nTopology platform which provided the mass properties for each geometry file. Overall, average expected weight of the designed specimens was to be 173.8 grams, with a range of approximately 11 grams between the lightest and heaviest. Then, based on the STL volumes provided by the build design software Materialise Magics, a density was calculated to determine variation between the nTop CAD model and the fabrication STL model. The further the density value is from 4.43 g/cc is an indication of how well the STL model matched the original CAD design. In this case,

it can be seen that the strut-type Octet and Kelvin were all slightly undersized in volume, supported by the increased density values. The TPMS lattices appear to have a higher accuracy for how well their STL mesh matched the implicit model. The average model volume was 38.7 mL with an average calculated density of 4.492 g/cc.

As for the EB-PBF fabricated dimensional accuracy, the beams showed slight deviations from nominal in length, width, and height. These were found to be consistent with the coefficient of thermal expansion (CTE) for Ti-6Al-4V, as shown in Table 5.13. Every dimension exhibited shrinkage from the nominal CAD geometry, indicating that Ti-6Al-4V parts to be produced by EB-PBF should have the models compensated for CTE at 600 °C [179], which is consistent with the chamber temperature. Therefore, 3D models should be scaled according to CTE such that they are printed relative to their expanded hot state, and subsequently shrink to nominal as they cool.

On average, the measured mass of the fabricated specimens, regardless of build orientation, was approximately 152 grams, representing a 12-13% reduction relative to the nominal design expectations. Similarly, volumetric assessments conducted via the Archimedes principle yielded values approximately 9% below the theoretical target, averaging around 35 mL. These deviations in both mass and volume are tell-tale signs of effective post-processing, specifically the successful removal of residual sintered powder from internal lattice regions. However, the discrepancy from nominal values resulted in an observed bulk density approximately 4% lower than the expected theoretical density for Ti-6Al-4V alloy, suggesting minor material loss or geometric deviations during the additive manufacturing process. More information on the mass and density measurements can be found in Appendix B.

Table 5.13. Dimensional values (average) measured from horizontally and vertically built lattice sandwich specimens compared with nominal (CAD) values, and calculated lengths based on the CTE at 600 °C.

	Nominal (CAD)	Horizontal		Vertical		lengths based on CTE @ 600 °C
		Averaged	% Error	Averaged	% Error	
Length (mm)	192	190.82	-0.61	190.80	-0.63	190.88
Width (mm)	48	47.98	-0.04	47.61	-0.81	47.72
Thickness (mm)	27.03	27.00	-0.10	27.01	-0.07	26.82
Mass (g)	173.84	152.47	-12.29	150.36	-13.51	
Volume (mL)	38.7	35.36	-8.66	35.00	-9.60	
Density (g/mL)	4.49	4.33	-3.71	4.31	-4.15	

5.4 4-POINT BEND FLEXURAL EXPERIMENTS

The four-point bend test is a critical mechanical evaluation method for titanium-based sandwich structures, as it simulates a realistic multi-axial loading environment representative of in-service conditions. This configuration subjects the lower face sheet to tensile stresses and the upper face sheet to compressive stresses, while the core experiences a complex stress state dominated by shear. The use of dual loading points generates a region of constant bending moment across the central span, thereby localizing failure to the structurally weakest zone and facilitating the identification of latent defects.

In conventional sandwich constructions, failure typically initiates through core shear, particularly along the interfacial bond lines between the face sheets and the core, which are often mechanically inferior. However, in the present study, the additively manufactured specimens feature a monolithic integration of face sheets and core, eliminating discrete bonding interfaces. Consequently, the failure mechanisms observed were distinct from traditional sandwich structures and were strongly influenced by the specific core geometries employed. Supporting data and information not presented in this section can be found in Appendix E.

5.4.1 *Beam Flexure Analysis*

Flexural behavior was characterized by measuring the applied force to the upper load points with a 250 kN load cell and comparing that data to the measure deflection of the beam at the lower center point by the deflectometer. The load versus displacement curves for the 55 horizontally built specimens, and 15 vertically built, all from the waste L-PBF Ti-6Al-4V powder, are shown in Figure 5.22. There were 4 repeated specimens of each design, except for the stress graded which only had three due to a premature build failure. Each plot compares the three grading methods with the different core lattice topologies. Overall, the results show a strong consistency for each configuration, which signifies sufficient build quality that was not defect driven.

All load-displacement curves exhibited the expected linear elastic behavior, characterized by a proportional increase in displacement with applied load during the initial loading phase. Across all core geometries, the uniformly graded specimens consistently initiated yielding prior to their density-graded counterparts, in agreement with predictions from finite element analysis. In every configuration, the stress-graded lattice cores demonstrated superior load-bearing capacity, with the TPMS Diamond structure achieving the highest peak load of 50 kN. However, this enhanced strength was accompanied by a reduction in flexural displacement, indicative of a more brittle failure response.

Notably, the stress-graded TPMS Gyroid and Diamond specimens exhibited abrupt failure immediately following peak load, whereas the remaining grading strategies and core geometries displayed more progressive damage evolution beyond maximum load. An additional trend observed was the earlier onset of failure in vertically oriented builds compared to horizontally fabricated specimens, highlighting anisotropic mechanical behavior as a function of build direction.

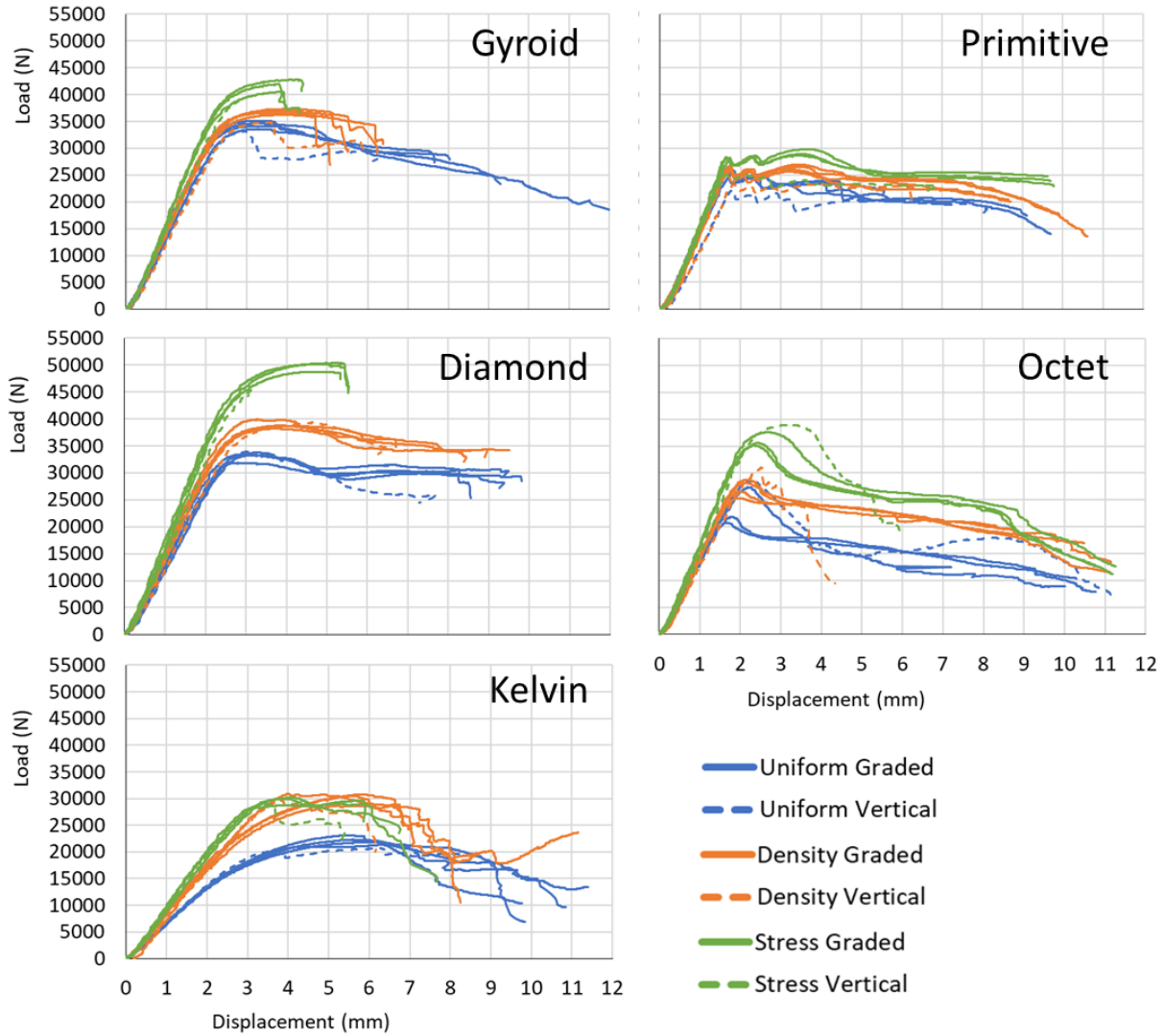


Figure 5.22. Load-Displacement results of the 4-point bend experiments for the Gyroid, Primitive, Diamond, Octet, and Kelvin sandwich cores and comparing the Uniform, Density, and Stress grading methods all at ~10% relative density.

The shape of the load–displacement curves exhibited a pronounced dependence on core geometry. TPMS-type Gyroid and Diamond structures typically attained peak load shortly after initial yielding, followed by a plateau phase as loading progressed toward failure. Uniformly graded variants of these cores demonstrated a slight post-yield decline in load-bearing capacity, indicative of localized structural degradation. In contrast, the strut-based Octet configuration

experienced a marked reduction in strength immediately following peak load, characteristic of core buckling behavior. Nevertheless, subsequent densification led to a stabilization of load capacity, approaching a plateau.

The Kelvin strut-type core, as anticipated, exhibited the lowest mechanical performance among the tested geometries. Its response was characterized by a gradual increase in load concurrent with disproportionately large deflections, revealing dominant failure mechanisms of progressive buckling and core crushing. TPMS Primitive cores showed minimal enhancement from grading strategies and displayed a distinctive bimodal peak in the load curve, succeeded by a displacement plateau. This behavior is attributable to a design deficiency that permitted extensive regions of the upper face sheet to undergo compressive buckling due to insufficient structural support.

Because load-displacement curves are geometry-specific, and describe how a structure behaves, they are less versatile for comparing different materials. While stress-strain curves are similar to load-displacement curves in that they depict the relationship between applied force and material deformation, they differ by their application and normalization. The stress-strain curves displayed in Figure 5.23 provide a better material-intrinsic representation by normalizing force by area (P/A), and displacement by original length. This approach is more applicable to various material systems like sandwich structures, and it provides for comparative properties such as the flexural modulus.

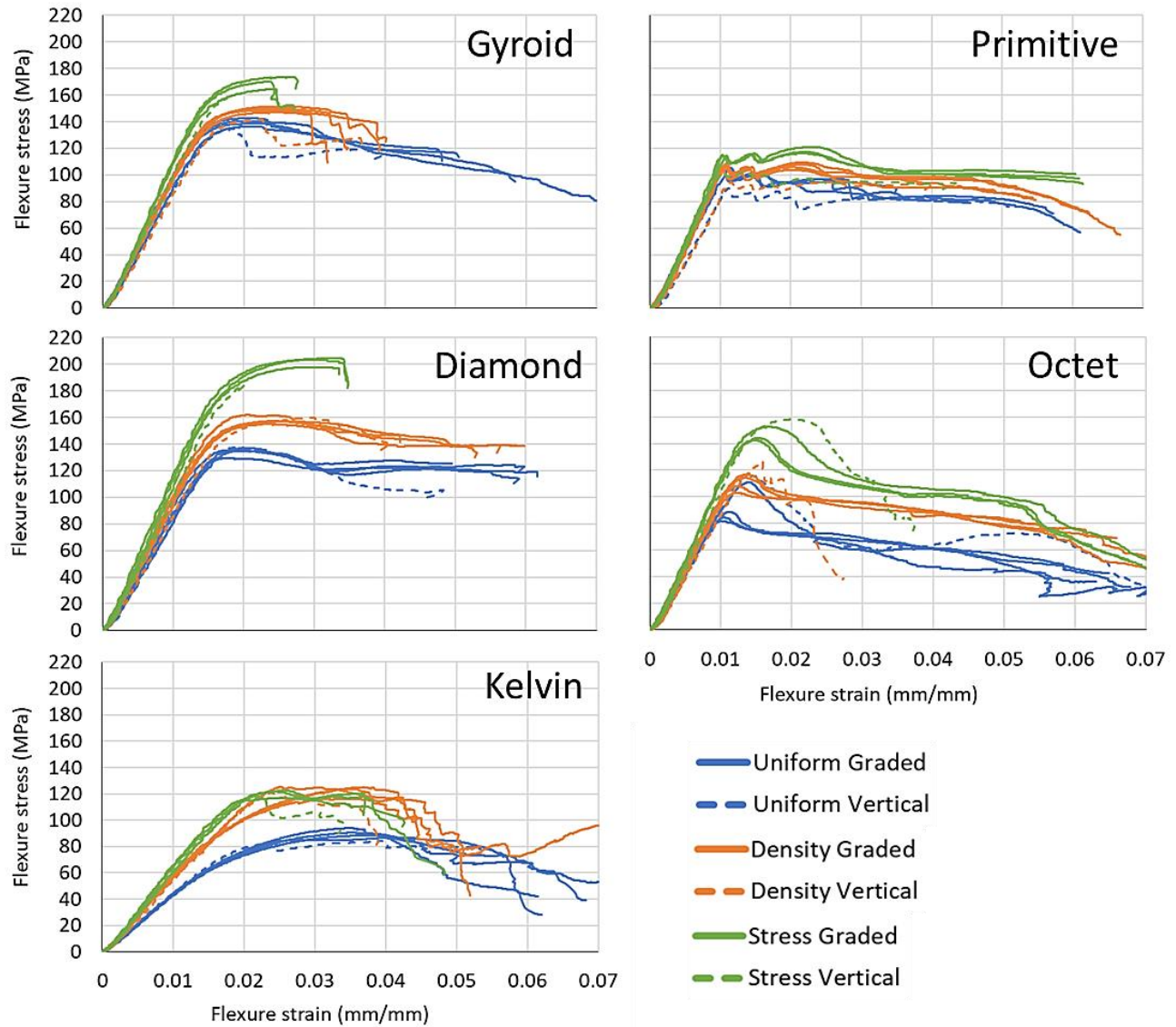


Figure 5.23. Stress-strain results of the 4-point bend experiments for the Gyroid, Primitive, Diamond, Octet, and Kelvin sandwich cores and comparing the Uniform, Density, and Stress grading methods all at ~10% relative density.

The flexural modulus, defined as the initial slope of the stress-strain curve under bending conditions, serves as a quantitative measure of structural stiffness across different grading strategies. Among the configurations evaluated, the stress-graded cores consistently exhibited the highest flexural stiffness, as evidenced by the steepest linear regions in their respective stress–

strain profiles across all core geometries. In contrast to uniaxial compression testing, where TPMS lattices typically demonstrate stretching-dominated deformation and strut-based lattices exhibit bending-dominated behavior, the mechanical response under four-point bending is more complex due to the simultaneous presence of tensile and compressive stresses. Nonetheless, the Octet lattice retains characteristics indicative of bending-dominated mechanics, whereas the TPMS Gyroid and Diamond structures display deformation patterns more consistent with stretching-dominated behavior.

For the Kelvin and the Primitive, these behavioral distinctions were less obvious. The low modulus and flatness of the uniform Kelvin curve would point to a bending-dominated behavior, although it showed the largest improvement with functional grading which seemingly made it behave more akin to a stretching-dominated lattice. In the case of the Primitive, it did not have a chance to display either since the core geometry allowed the upper face sheet to buckle. Post-buckling, the structure began to fold, however the density and stress graded cores still retained their stiffness which led to a slight increase in strength beyond the flexural strain of 0.02.

The mechanical response of vertically fabricated specimens revealed a pronounced deviation from their horizontally built counterparts. In nearly all cases, excluding the Octet configuration, vertically oriented cores exhibited earlier onset of structural failure and reduced overall load-bearing capacity. Notably, across all grading strategies, the vertically built Octet specimens demonstrated higher peak stress and strain values compared to their horizontal equivalents. However, once failure was initiated, these structures underwent rapid collapse, with limited post-yield stability. The vertically fabricated, stress-graded Gyroid and Diamond cores displayed markedly lower flexural modulus, followed by abrupt failure immediately upon yielding. This

behavior was characterized by the absence of a post-yield plateau, and indicative of a brittle failure mode with diminished energy absorption capacity under flexural loading.

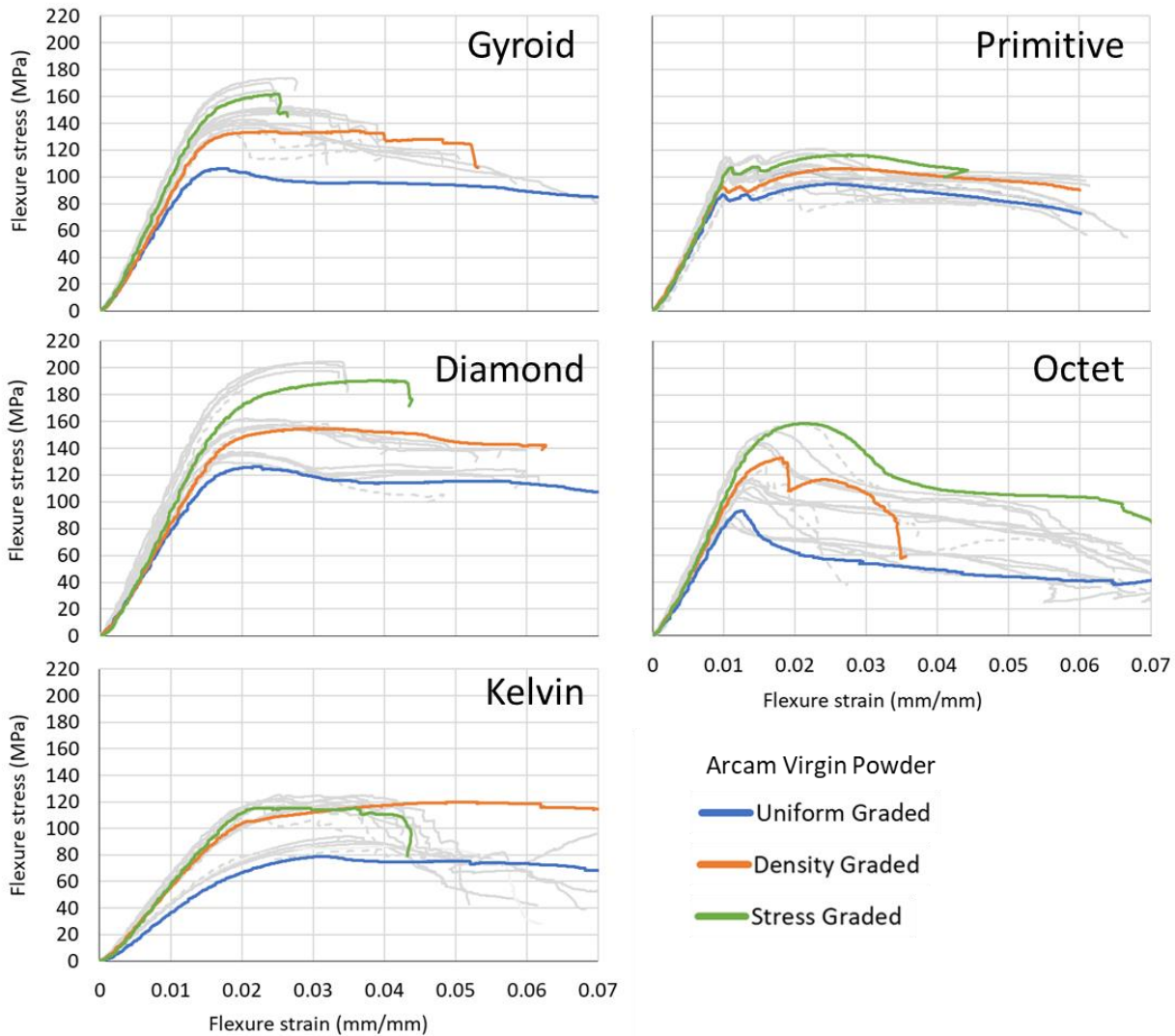


Figure 5.24. Stress-strain results from specimens produced using Arcam Virgin powder compared to those manufactured from the waste L-PBF powder (grayed out).

In comparison with the waste L-PBF powder, the same 15 specimens were produced using virgin EB-PBF powder, procured from the machine manufacturer, Arcam. The flexural stress-strain behavior of the Arcam virgin powder specimens is depicted in Figure 5.24, overlaid with the Figure 5.23 results in grey. For the TPMS Gyroid and Diamond, the linear regions have a

slightly lower slope (i.e. reduced stiffness), and the curves peak at lower stresses. The Primitive and Kelvin do not display as noticeable of a difference, though the density graded Kelvin showed a remarkable robustness as it held its strength nearly to the end of the test. In the case of the Octet, the virgin powder specimens outperformed those built with the waste L-PBF powder, demonstrating higher overall strength, particularly with the stress graded core.

In summary, the beam flexural analysis revealed distinct mechanical behaviors across core geometries and grading strategies, with stress-graded TPMS structures consistently demonstrating superior stiffness and peak load capacity. The influence of build orientation was particularly pronounced, underscoring the anisotropic nature of EB-PBF produced lattices. While TPMS cores generally exhibited stretching-dominated responses and strut-based designs leaned toward bending-dominated failure, the complexity of four-point bending introduced hybrid deformation modes that merit further investigation. To elucidate the underlying failure mechanisms and validate these structural trends, a detailed fractographic examination was conducted. The following section presents an assessment of fracture surfaces to correlate observed mechanical responses with localized damage evolution.

5.4.2 *Fractography*

To complement the mechanical characterization presented in the preceding section, fractographic analysis was conducted to investigate the failure mechanisms. By examining the fracture behavior and fracture surfaces of tested specimens, insights were gained into the dominant modes of deformation and damage propagation associated with different core geometries, grading strategies, and build orientations. This qualitative assessment provides critical context for interpreting the observed variations in flexural performance, particularly in relation to localized buckling, delamination, and brittle fracture phenomena. The following results elucidate how structural

topology influence failure morphology, thereby informing future design optimization for enhanced mechanical resilience.

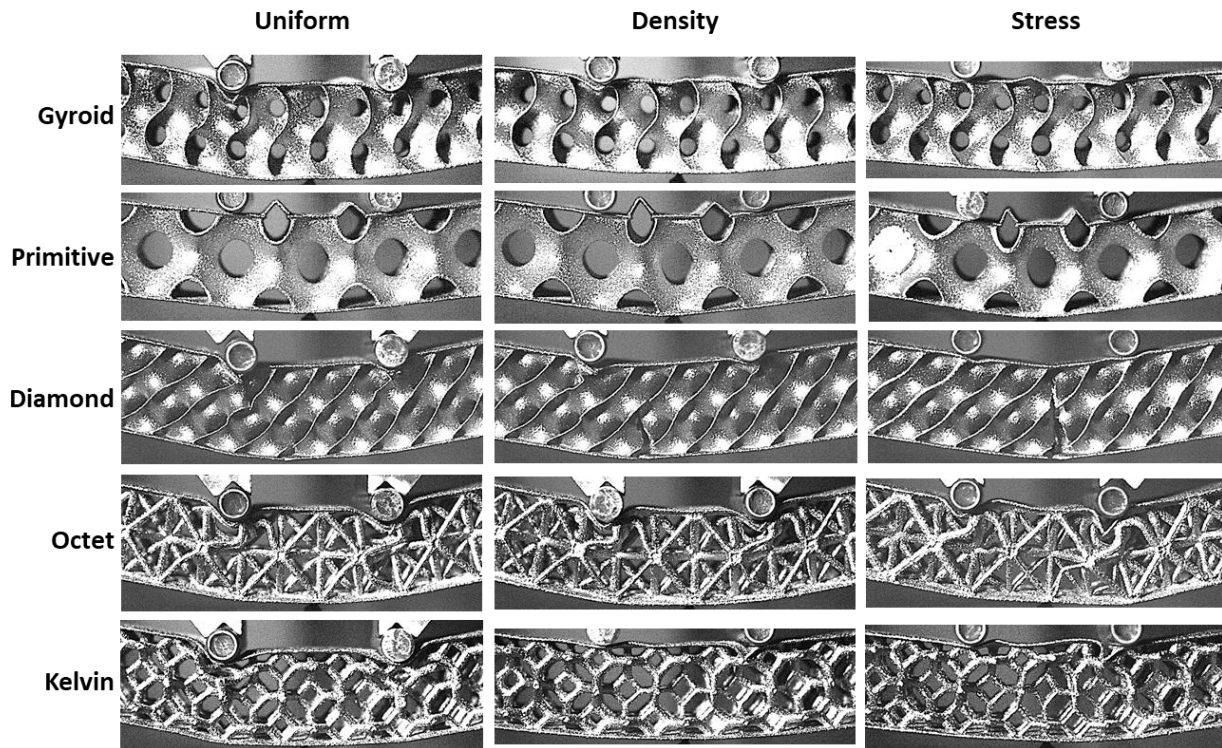


Figure 5.25. Post 4-point bend experiment images depicting the failure modes of the various sandwich beam specimens.

In addition to the quantitative insights derived from stress-strain analysis, visual examination of the post-failure specimens offered valuable context regarding the response of the fifteen distinct sandwich panel configurations. Figure 5.25 presents representative fracture images for each design, organized to facilitate comparative evaluation. The columns align specimens by core geometry under a consistent thickness grading strategy, while the rows delineate the effects of the three grading methodologies. A discernible trend emerges across the horizontal axis, where damage progressively diminishes from uniformly graded to density-graded, and ultimately to stress-graded structures. Vertically, the images reveal that TPMS-based cores exhibited greater

resistance to core buckling relative to their strut-based counterparts, underscoring the influence of topology on failure mode and structural integrity.

For the Gyroid and Diamond core configurations, visual inspection revealed pronounced damage localized beneath the load application points in specimens with uniform thickness grading. This observation underscores the efficacy of both density and stress grading strategies in mitigating structural compliance and enhancing load distribution. In the region between the load points, subjected primarily to compressive forces, localized core compaction was observed, which induced buckling of the upper face sheet. The relative vertical displacement of the upper loading pins, as captured in the images, reinforces that the uniformly graded cores experienced greater flexural strain compared to their density-graded counterparts, with stress-graded specimens exhibiting the least strain.

These qualitative findings are in strong agreement with the preceding stress-strain curves. Although fracture features were not distinctly visible in the captured images, failure in the density- and stress-graded specimens consistently initiated in tension at the lower face sheet between the load points, with crack propagation occurring vertically through the core structure. This failure mode was similarly observed in the majority of uniformly graded specimens, with the exception of a single Gyroid sample that exceeded the strain capacity of the test apparatus prior to catastrophic failure.

The TPMS Primitive fracture images presented in Figure 5.25 provide valuable context for interpreting the corresponding stress-strain behavior. In each case, the spatial arrangement of the core geometry relative to the loading direction resulted in localized buckling of the upper and lower face sheets between the load application points, particularly in regions lacking sufficient core support. The characteristic double peak observed at the termination of the linear elastic region

in the stress-strain curves is attributed to sequential buckling events, initially in one unsupported section of the face sheet, followed closely by the second. Following the onset of compressive buckling in the upper face sheet, the residual flexural resistance was governed solely by the stiffness of the core structure.

Given that the face sheets were geometrically and materially identical across all Primitive configurations, peak load values remained relatively consistent. However, both density- and stress-graded cores demonstrated a modest delay in the onset of buckling, attributable to more efficient material distribution and localized reinforcement. A noteworthy observation for the Primitive core is its ability to sustain strain beyond face sheet failure without significant structural compromise to the core itself. The stress response remained relatively stable post-buckling, and no tensile failure was observed in the lower face sheet. Ultimately, testing was terminated prematurely due to excessive lateral loading on the apparatus, which was induced by the loss of structural continuity in the upper face sheet.

The Octet lattice core exhibited buckling behavior characteristic of strut-based architectures under flexural loading. Because strut diameter was determined using a linear static FEA, it did not account for nonlinear buckling phenomena. As a result, all grading strategies, including density and stress-based approaches, demonstrated similar failure modes to the uniformly graded configuration. In the uniform core specimens, the load application pins induced deeper penetration into the panel, accompanied by pronounced localized deformation. In contrast, the stress-graded cores, and to a lesser extent the density-graded variants, exhibited improved resistance to localized buckling at the load points, thereby limiting both penetration depth and core deformation. However, the presence of thinner struts in the lower regions of the graded panels highlights a deficiency in structural sizing with respect to column buckling resistance, alluding that future

design iterations should incorporate nonlinear stability analyses to better capture and mitigate these failure mechanisms.

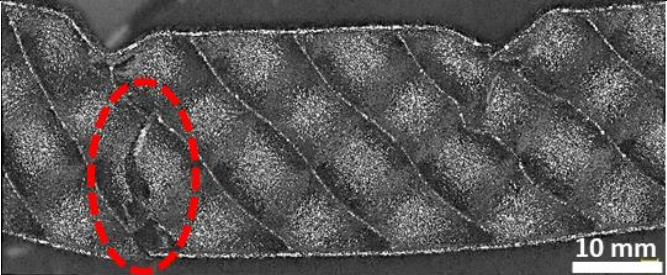
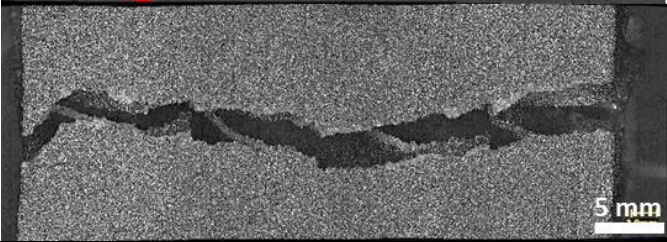
The images presented in Figure 5.25 exposes the sensitivity of the Octet lattice configuration to precise alignment within the experimental test setup. During compressive loading, initial buckling consistently occurred in the struts directly beneath the load application points. The timing of this buckling onset was found to be highly dependent on the degree of alignment between the struts and the load pins; even minor misalignments introduced additional bending moments, thereby accelerating structural failure in those regions. Despite the extensive core buckling and localized damage inherent to the Octet architecture, the panel did not exhibit tensile failure in the lower face sheet. This absence of tensile rupture suggests that the structural integrity of the face sheet remained sufficient to prevent catastrophic loss of load-bearing capacity, even under conditions of significant core degradation.

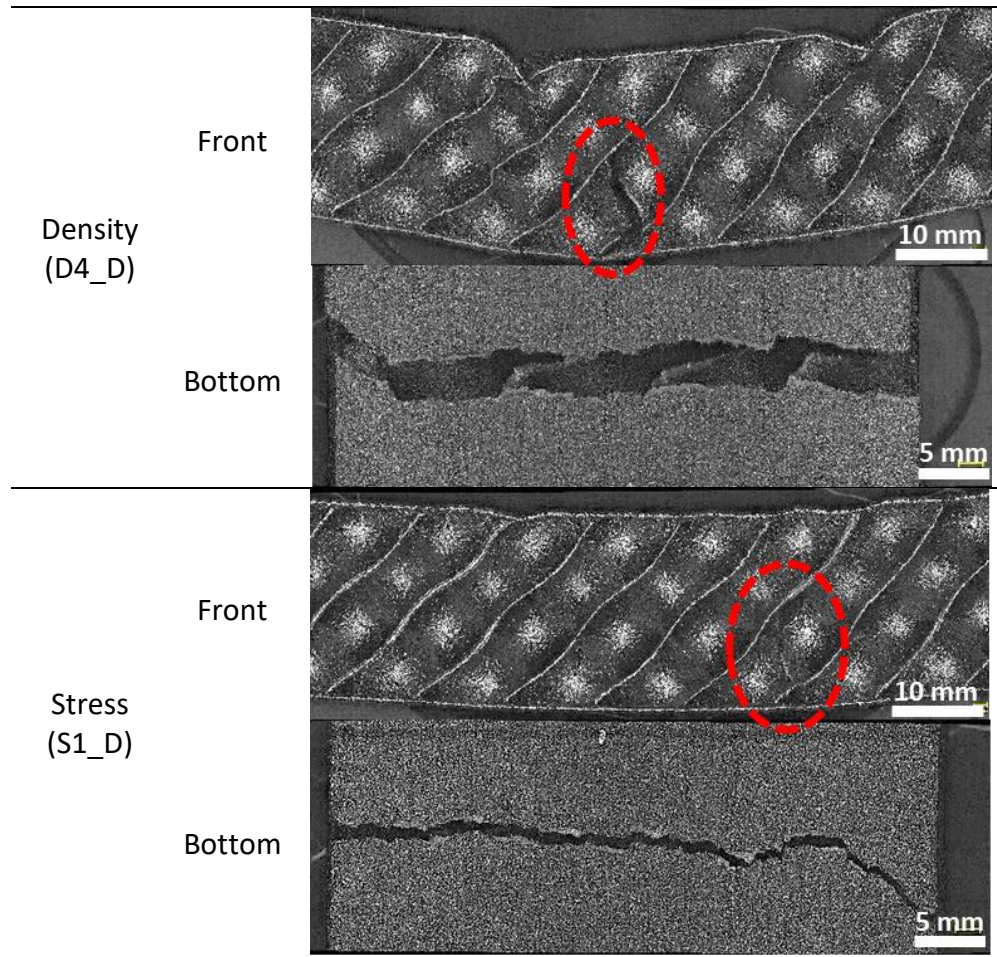
For the Kelvin core configuration, minimal variation was observed in the stress-strain response between the density-graded and stress-graded topologies, as evidenced by the failure modes depicted in Figure 5.25. The uniformly graded core exhibited more pronounced damage in the vicinity of the load application points. During deformation, tensile stresses developed in the upper face sheet, resulting in its relatively planar and undeformed appearance. In contrast, both density-graded and stress-graded panels displayed buckling and surface undulations in the upper face sheet, symptomatic of more distributed strain localization. Although the overall core geometries appeared structurally intact in the post-failure imagery, numerous individual struts were fractured or disconnected, compromising internal connectivity. At elevated flexural strain levels, progressive core crushing was accompanied by audible high-frequency “ting” events, which

correlated with abrupt stress drops in the mechanical response curves, indicating discrete strut failures and sudden loss of load transfer pathways within the lattice network.

Because the TPMS Diamond specimens failed at the lower face sheet, a more detailed examination of the fracture surfaces was performed. Table 5.14 presents a comparative optical analysis of three representative Diamond specimens, each corresponding to one of the three grading strategies: uniform, density, and stress-based. These specimens are symbolic of the broader set of Diamond configurations, which exhibited consistent structural behavior across design groups. Fracture localization was confined to the region between the two load application points; however, the precise crack initiation sites varied, with some specimens failing near mid-span and others closer to the left or right loading positions. This variability is likely attributable to minor inconsistencies in experimental setup, as beam centering on the test fixture was performed through visual alignment rather than precision tooling.

Table 5.14. Post-failure optical images of the Diamond sandwich structure with fracture locations identified.

Specimen	View	Optical Image
Uniform (U4_D)	Front	
	Bottom	



Regardless of the curvature inherent to the TPMS Diamond topology or the specific grading methodology employed, tensile failure of the lower face sheet consistently initiated a vertical crack that propagated through the core structure. Observable effects of lattice orientation were evident on the surface of the bottom face sheet, where the fracture path remained predominantly linear until exhibiting a pronounced deviation of approximately 45 degrees near one edge. This indicates that crack initiation likely occurred at the edge adjacent to a core interface, where localized stress concentrations exceeded the material's tensile strength. The crack subsequently followed the geometric contour of the core until encountering a discontinuity in the lattice orientation, at which point it resumed a relatively linear trajectory through the remaining core volume.

Localized buckling at the load application interfaces was evident in both the uniform and density-graded Diamond sandwich structures, and this mechanical instability was associated with diminished load-bearing capacity and post-peak softening behavior. Despite the occurrence of core crushing in these regions, the damage did not propagate throughout the structure nor significantly influence the ultimate fracture mode. The similarity in deformation patterns provides that the topology optimization approach based on density grading was insufficient in accommodating the specific loading conditions. In contrast, specimens utilizing stress-based grading exhibited minimal local buckling and negligible core crushing prior to failure, thereby maintaining structural integrity and avoiding post-peak softening. The absence of localized damage underscores the efficacy of stress-driven grading as a robust strategy for functionally tailoring lattice core thicknesses in sandwich structures. Extending this methodology to include optimization of face sheet thicknesses, treating the sandwich system as an integrated whole, could potentially yield further enhancements in mechanical performance.

Scanning Electron Microscopy (SEM) analysis of the fracture surface from a TPMS Diamond specimen revealed characteristic features indicative of tensile failure in a ductile alloy, particularly Ti-6Al-4V. The predominant morphological evidence consisted of a dimpled surface texture, as illustrated in Figure 5.26a, which is consistent with microvoid nucleation, growth, and coalescence during plastic deformation. These dimples represent the final stages of void evolution as the material undergoes tensile separation. Additionally, localized necking was observed along the crack path, although its identification was impeded by the inherent surface roughness at the microscale.

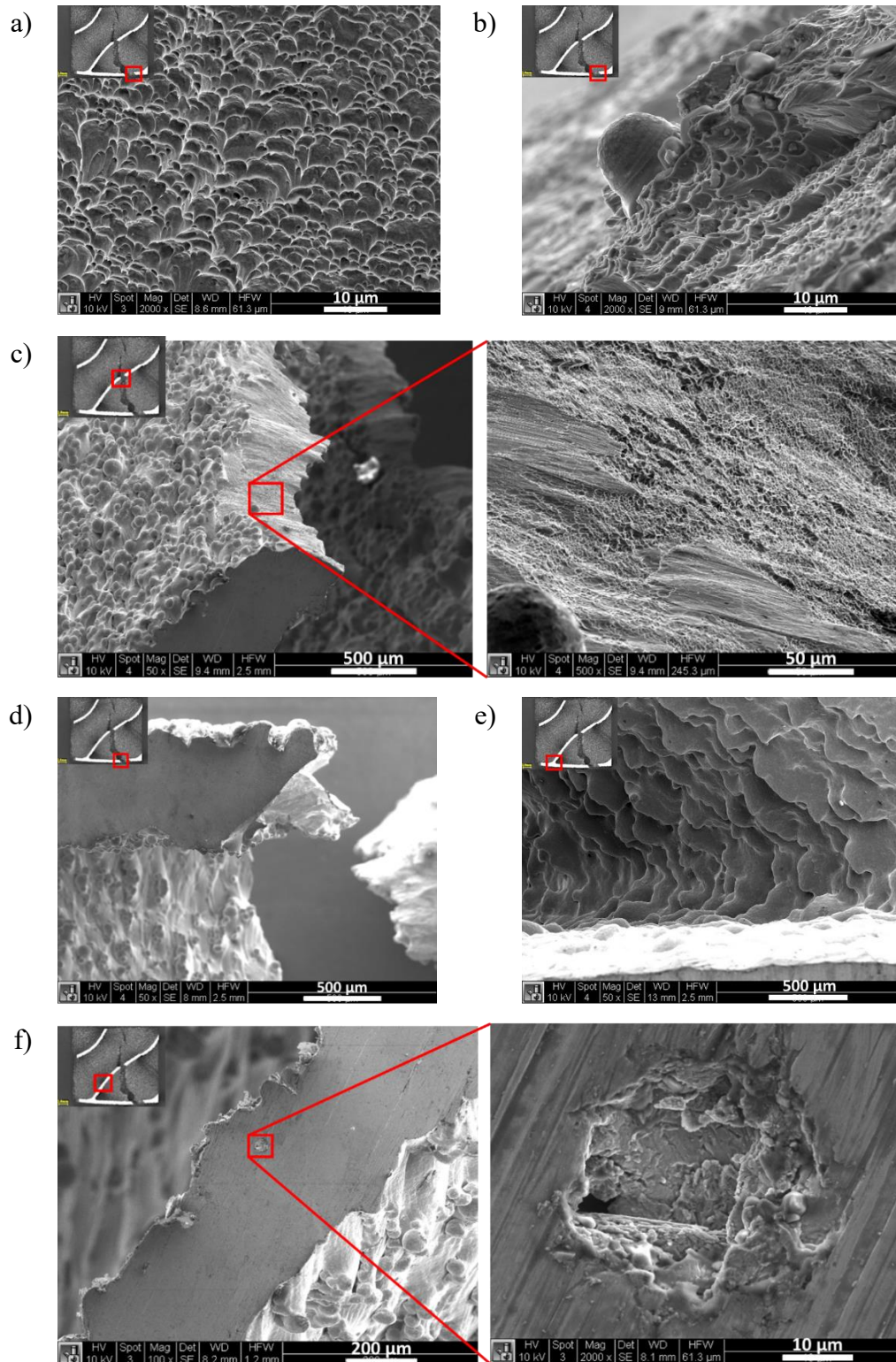


Figure 5.26. SEM images from specimens S1_D and SV_D displaying a) high resolution of typical fracture surface, b) potential defect in fracture surface, c) broad view of fracture surface with closeup, d) side view of fracture, e) blend radius interface, f) pore with closeup.

A comprehensive examination of the fracture surfaces revealed a predominantly defect-free morphology, evidence of high material integrity, and that mechanical performance was not significantly influenced by manufacturing-induced defects (Figure 5.26c). Isolated instances of potential lack-of-fusion voids were observed, characterized by the presence of smooth, partially melted powder particles protruding from the fracture surface, as shown in Figure 5.26b. Variations in local surface roughness appeared to influence the elevation profile along the crack path. Consequently, the interplay between thickness irregularities and the prevalence of geometrically sharp crack initiation sites likely contributed to localized deviations in crack trajectory, resulting in micro-scale meandering behavior during fracture propagation.

As displayed in Figure 5.26d, which presents a lateral view of the sectioned fracture surface, failure was observed to propagate along a shear plane oriented approximately 45° relative to the principal tensile stress and loading axis. This shear-dominated fracture mode is further corroborated by the presence of elongated and directionally stretched dimples on the fracture surface, emblematic of plastic deformation under shear stress. Surface roughness variations were noted across the material thickness; however, the overall thickness was deemed sufficient to ensure that sandwich structure performance was not significantly influenced by these topographical irregularities. Subsurface inspection revealed isolated lack-of-fusion voids embedded within the bulk material (e.g., Figure 5.26f), indicating minor process-induced defects. The SEM micrograph in Figure 5.26e demonstrates the fidelity with which intricate design features, such as the core-to-face sheet blend radius, were realized by the EB-PBF fabrication method. Nevertheless, the observed surface resolution implies that a blend radius greater than 0.8 mm may be necessary to ensure clearer geometric definition and improved manufacturability.

In summary, the fractographic examination revealed distinct failure mechanisms across the various sandwich core configurations, with features such as dimple rupture, cleavage facets, and intergranular fracture correlating closely with the observed mechanical performance. The presence of unmelted particles, microvoid coalescence, and column buckling in select regions further underscored the influence of lattice geometry on fracture behavior. These findings not only validate the mechanical test results but also highlight the need for a deeper understanding of the underlying microstructural characteristics that govern crack initiation and propagation. To this end, the following section presents a microstructure analysis aimed at revealing the phase composition, grain morphology, and defect distribution that contributed to the observed fracture modes.

5.4.3 *Microstructure Analysis*

To complement the mechanical and fractographic evaluations, microstructural analysis was conducted to investigate the internal morphology and phase distribution of the additively manufactured Ti-6Al-4V sandwich structures. This section presents the results obtained through optical microscopy, with particular emphasis on grain structure, orientation, phase composition, and defect characterization across different face sheet and core characteristics. The microstructural features are critical for interpreting the mechanical performance, as they directly influence properties such as strength, ductility, and fracture behavior. Variations in build orientation, thermal history, and lattice topology are examined to illuminate their role in shaping the microstructural evolution and overall material integrity.

Figure 5.27 presents the microstructural characterization of stress-graded TPMS Diamond specimens fabricated via horizontal and vertical build orientations, with cross-sectional samples extracted adjacent to the fracture sites. The observed microstructure is consistent with the typical

Ti-6Al-4V ($\alpha + \beta$) Widmanstätten morphology, comprising prior- β grains populated by intragranular α and β phases. Notably, the peripheral regions of the specimens exhibit pronounced surface irregularities, characterized by sharp valleys and asperities that are likely to compromise fatigue resistance. As anticipated for components produced by the EB-PBF process, the downward-facing surfaces, those oriented opposite to the build direction, demonstrate significantly greater roughness compared to their upward-facing counterparts in horizontally built samples. Additionally, the presence of microscopic porosity is evident throughout the micrographs, with a markedly higher pore density observed in vertically built specimens relative to those fabricated horizontally.

Figure 5.27a presents a comparative analysis of the bottom face sheet microstructures for specimens fabricated in horizontal and vertical build orientations, highlighting distinct differences in grain morphology. In the horizontally built samples, columnar prior- β grains are observed extending through successive layers along the build direction, indicative of directional solidification driven by thermal gradients during layer-by-layer deposition. In contrast, the vertically built specimens exhibit a more equiaxed grain structure, which results from the cross-sectional plane being oriented perpendicular to the build direction, effectively providing a top-down view of the columnar grain growth. This orientation-dependent perspective accounts for the apparent morphological differences between the two samples. In the horizontal configuration, grain columns initiate at the part boundary and propagate inward at an angle, reflecting the influence of localized remelting and solidification dynamics. The observed grain inclination corresponds to the solidification front, where the outer regions of the melt pool experience accelerated cooling rates, thereby dictating the directionality of grain growth.

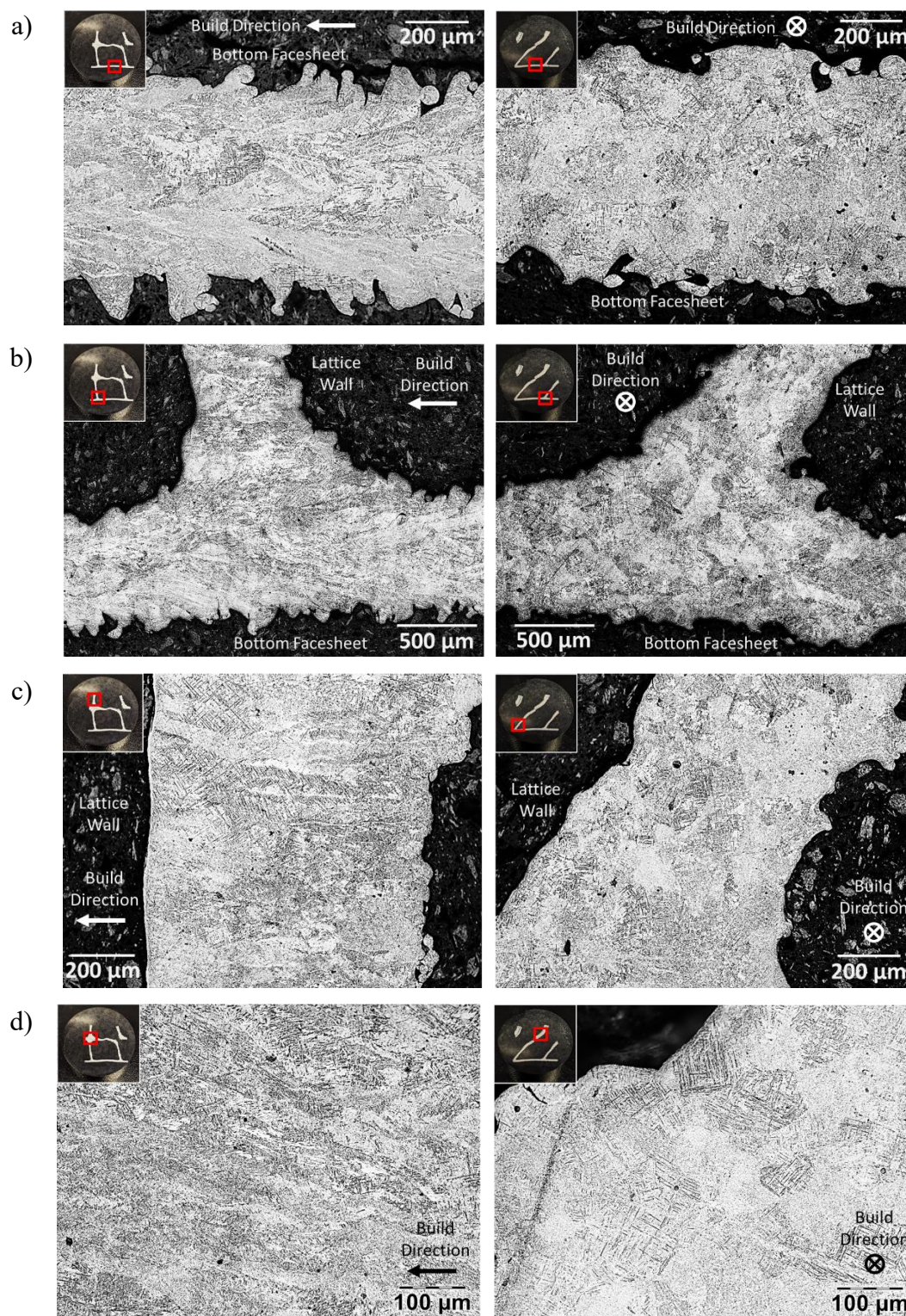


Figure 5.27. Microstructure comparison of Ti-6Al-4V horizontally and vertically built Diamond lattice sandwich structures at the a) bottom face sheet, b) face sheet-core interface, c) lattice wall, and d) higher resolution of typical microstructure texture.

Figure 5.27b illustrates the microstructural characteristics at the interface between the face sheet and the core region. The distinction between columnar and equiaxed grain morphologies remains evident, with the vertically fabricated specimen exhibiting a consistent microstructure across the transition zone, indicating minimal disruption to grain morphology. In contrast, the horizontally fabricated specimen displays columnar grains aligned with the face sheet direction that persist unchanged across the interface, as anticipated based on the build orientation.

This continuity in grain alignment results in the lattice wall microstructure being oriented through the material thickness. In regions where the lattice walls are relatively thin, columnar grains were observed traversing the entire thickness from one surface to the opposite side, displaying uninterrupted grain growth across the wall section. Figure 5.27c provides a full-thickness view of the lattice walls, enabling a comparative assessment of microstructural features between the two build orientations, while Figure 5.27d offers a more detailed depiction of the representative grain structures.

The microstructural observations presented throughout this section underscore the critical role of build orientation in governing grain morphology and alignment within EB-PBF fabricated components. Vertically oriented builds tend to produce equiaxed grains with uniformity across transition zones, while horizontally oriented builds favor the development of columnar grains that persist through structural interfaces and align with the build direction. This directional grain continuity, particularly evident in thin lattice walls, provides a strong correlation between thermal gradients and solidification behavior. The comparative analyses in Figure 5.27a through Figure 5.27d provide compelling evidence of how orientation-dependent solidification dynamics influence microstructural evolution, with direct implications for mechanical anisotropy and performance in additively manufactured sandwich structures.

5.4.4 *Final Material Characterization*

It is well established that repeated reuse of Ti-6Al-4V powder in EB-PBF processing leads to progressive oxidation of powder particles. The resultant oxygen uptake enhances the strength of the consolidated material but typically reduces its ductility due to interstitial hardening effects. In the fabrication of both horizontally and vertically oriented sandwich beam specimens, waste L-PBF powder was employed across eleven successive builds. As powder losses occurred during processing, additional material was intermittently introduced and blended with the existing feedstock, resulting in a heterogeneous powder history. To evaluate the cumulative impact of powder reuse on mechanical properties, a twelfth build consisting of tensile coupons was conducted using the final state of the waste L-PBF powder. These specimens were strategically positioned to align with the locations of the previously fabricated lattice beams, as illustrated in Figure 5.28, enabling assessment of potential spatial variations in mechanical performance within the build chamber.

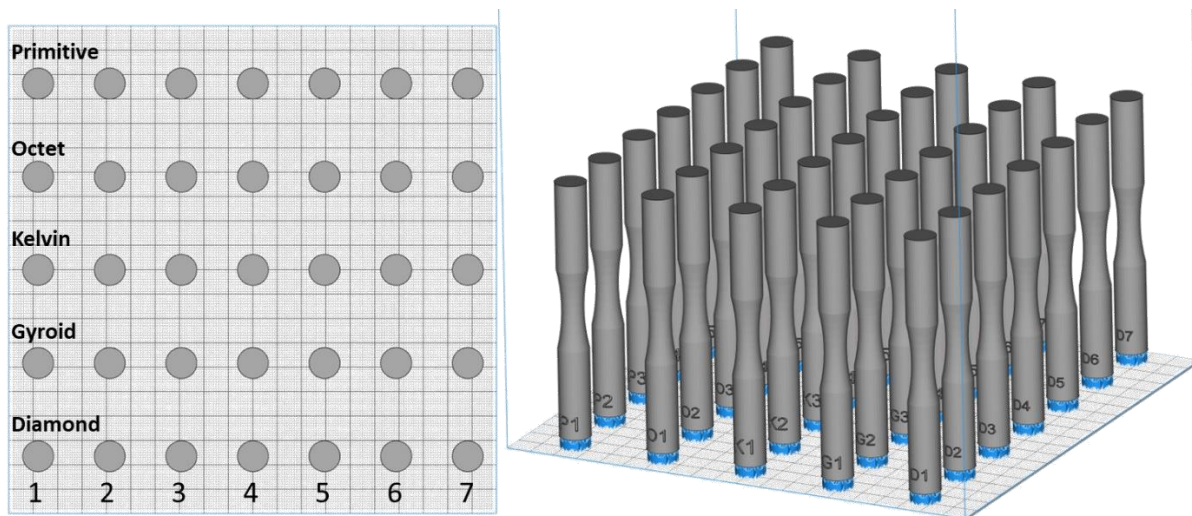


Figure 5.28. Final used L-PBF powder and virgin EB-PBF powder tensile specimen build arrangement.

Two sets of tensile specimens were fabricated to evaluate the mechanical implications of powder reuse. One set utilized the waste L-PBF powder, while the other employed virgin Arcam EB-PBF powder that had not yet been used in the fabrication of sandwich beam structures. Tensile experiments were conducted, and the resulting mechanical behavior is presented in Figure 5.29. The data reveal a pronounced divergence in performance between the two powder conditions. Specimens produced from virgin powder exhibited lower ultimate tensile strength but retained significantly higher ductility. In contrast, those fabricated from reused powder demonstrated an approximate 100 MPa increase in ultimate tensile stress, accompanied by a substantial reduction in elongation to failure, approximately half that of the virgin powder specimens. These results are contextualized against the baseline mechanical properties of AMS 4911, a standard Ti-6Al-4V sheet material commonly employed in the aerospace industry.

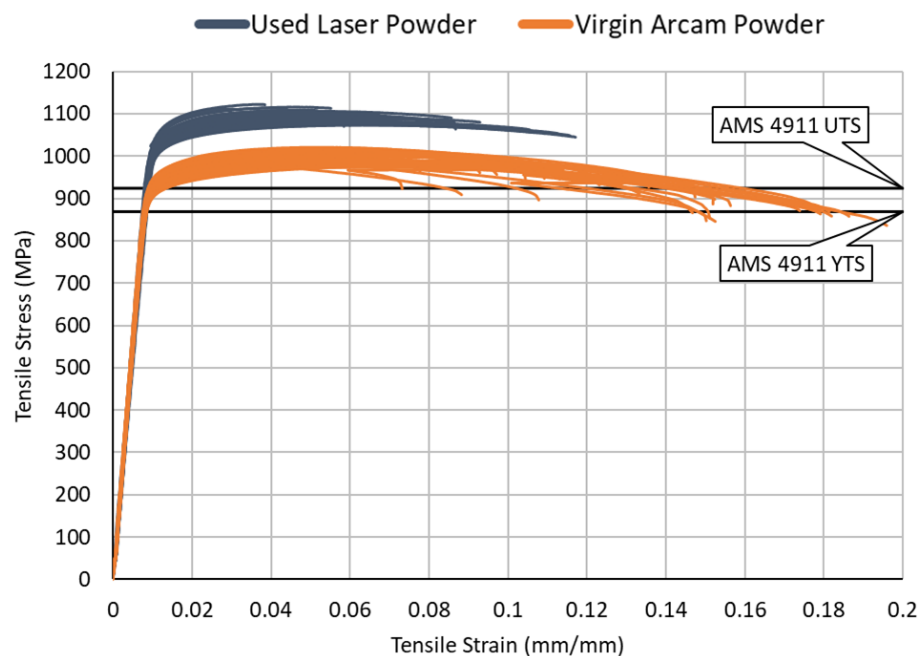


Figure 5.29. Final used L-PBF (“Laser”) powder tensile behavior alongside virgin EB-PBF (“Arcam”) powder.

Figure 5.30 and Figure 5.31 present the spatial distribution of tensile properties for specimens fabricated using both used L-PBF and virgin EB-PBF powder, mapped according to their grid positions within the build chamber as defined in Figure 5.28. The visualizations employ a color scheme to illustrate deviations in mechanical performance. On the left-hand grids, color intensity corresponds to variation from the designated “Initial” reference value, which was the baseline tensile property obtained from testing of the as-received waste L-PBF powder prior to its use in the current study. In this context, blue shading indicates values exceeding the initial benchmark, whereas red denotes values below it, with each property normalized to its respective scale. The right-hand grids depict the percentage deviation from the mean value of the experimental group, where green signifies performance above the group average and red shows performance below it. This spatial mapping facilitates assessment of chamber-specific influences on mechanical behavior and highlights potential heterogeneity from powder condition and build location.

In general, the final values of the used L-PBF powder shown in Figure 5.30 exhibit higher strength and lower ductility. The material required an average of 8 kN more load to fail than before the sandwich specimen builds, and average yield and ultimate tensile stress values increased by 113 MPa and 102 MPa, respectively. Elastic modulus increased from 104.7 GPa to 111.1 GPa, meanwhile elongation plummeted from 18% to 6.4%, and strain at maximum load dropped from 0.065 to 0.048 mm/mm. Spatially, those specimens at the central region of the build area, which correspond to the Octet, Kelvin, and Gyroid mid-spans, had properties closer to the initial values. The center of the build area was also where the initial tensile specimens were located. Therefore, the location of the Diamond structure in the build chamber may have been a contributing factor in its higher strength performance. Because the upper face sheets buckled on the Primitive lattice beams, it is unlikely its performance was affected by its print location.

Maximum Load (kN)															
				Group Avg	Initial	% difference									
				69.7	61.5	13.4		Row Avg	% difference from group average value						
68.4	69.8	70.0	69.3	70.1	70.4	70.7		69.8	-1.9	0.2	0.5	-0.5	0.6	1.0	1.5
70.0	70.0	No Data	69.2	69.4	70.1	70.7		69.9	0.5	0.4	No Data	-0.7	-0.4	0.6	1.5
71.0	69.5	68.8	68.3	68.6	69.4	70.6		69.5	1.9	-0.3	-1.3	-2.0	-1.5	-0.4	1.2
70.4	68.9	68.5	68.3	68.6	69.3	70.1		69.2	1.0	-1.1	-1.7	-2.0	-1.6	-0.6	0.6
71.4	70.0	69.7	69.5	69.7	69.8	70.8		70.1	2.5	0.4	0.0	-0.3	0.0	0.2	1.6

Tensile stress at Max Load (MPa)															
				Group Avg	Initial	% difference									
				1095	993	10.3		Row Avg	% difference from group average value						
1075	1097	1101	1090	1102	1107	1112		1098	-1.9	0.2	0.5	-0.5	0.6	1.0	1.5
1100	1100	No Data	1088	1091	1102	1111		1099	0.5	0.4	No Data	-0.7	-0.4	0.6	1.5
1116	1092	1082	1073	1079	1091	1109		1092	1.9	-0.3	-1.3	-2.0	-1.5	-0.4	1.2
1106	1083	1077	1074	1078	1089	1102		1087	1.0	-1.1	-1.7	-2.0	-1.6	-0.6	0.6
1122	1100	1096	1092	1095	1097	1113		1102	2.5	0.4	0.0	-0.3	0.0	0.2	1.6

Modulus (GPa)															
				Group Avg	Initial	% difference									
				111.1	104.7	6.1		Row Avg	% difference from group average value						
109.8	116.1	112.8	111.2	111.1	112.8	112.1		112.3	-1.2	4.5	1.5	0.0	0.0	1.5	0.9
105.2	116.6	No Data	116.3	110.1	120.5	113.0		113.6	-5.4	4.9	No Data	4.7	-1.0	8.4	1.7
112.0	109.7	109.1	110.4	106.5	111.2	107.4		109.5	0.8	-1.3	-1.8	-0.7	-4.1	0.0	-3.3
110.7	108.7	105.0	108.5	109.6	110.7	114.0		109.6	-0.4	-2.2	-5.6	-2.3	-1.4	-0.4	2.5
114.1	111.9	109.4	110.1	109.1	110.8	112.2		111.1	2.6	0.7	-1.5	-1.0	-1.8	-0.3	0.9

Elongation (%)															
				Group Avg	Initial	% difference									
				0.064	0.18	-64.6		Row Avg	% difference from group average value						
0.024	0.037	0.050	0.034	0.051	0.044	0.029		0.039	-62.5	-41.6	-21.0	-46.2	-19.3	-30.7	-54.1
0.032	0.078	No Data	0.061	0.093	0.067	0.032		0.061	-50.2	22.9	No Data	-4.3	45.9	5.8	-49.5
0.055	0.060	0.105	0.106	0.103	0.052	0.052		0.076	-13.3	-5.8	65.3	66.6	61.8	-18.7	-18.8
0.086	0.086	0.116	0.117	0.113	0.050	0.054		0.089	34.6	35.1	81.8	84.0	76.8	-21.2	-15.6
0.038	0.064	0.069	0.058	0.064	0.040	0.044		0.054	-39.6	0.3	7.7	-8.2	-0.1	-37.3	-30.5

Tensile Yield stress (0.2% Offset) (MPa)															
				Group Avg	Initial	% difference									
				1026.2	913.4	12.3		Row Avg	% difference from group average value						
1020	1029	1028	1024	1031	1036	1052		1031	-0.6	0.3	0.2	-0.2	0.5	0.9	2.5
1035	1043	No Data	1016	1022	1032	1046		1032	0.9	1.7	No Data	-0.9	-0.4	0.6	1.9
1042	1022	1013	1000	1007	1022	1035		1020	1.5	-0.4	-1.3	-2.5	-1.8	-0.4	0.9
1032	1011	1008	1004	1010	1017	1031		1016	0.6	-1.5	-1.7	-2.2	-1.6	-0.8	0.5
1049	1028	1025	1024	1024	1027	1043		1031	2.3	0.2	-0.1	-0.3	-0.2	0.1	1.6

Tensile strain at Max Load (mm/mm)															
				Group Avg	Initial	% difference									
				0.048	0.065	-26.6		Row Avg	% difference from group average value						
0.023	0.033	0.047	0.032	0.047	0.042	0.027		0.036	-52.3	-30.6	-2.1	-33.5	-0.6	-12.6	-42.7
0.031	0.053	No Data	0.050	0.063	0.050	0.030		0.046	-35.8	10.3	No Data	4.4	31.5	4.4	-36.7
0.045	0.049	0.063	0.063	0.062	0.049	0.047		0.054	-5.4	2.7	33.0	32.1	30.5	3.0	-2.1
0.050	0.057	0.063	0.059	0.062	0.052	0.046		0.056	5.3	20.3	31.4	24.2	30.7	9.8	-3.0
0.037	0.052	0.057	0.053	0.049	0.039	0.040		0.047	-22.4	8.4	19.5	10.8	3.2	-19.1	-16.9

Figure 5.30. Final used L-PBF powder tensile properties (colors scales relative to the individual table and labeled “Initial” value).

The virgin EB-PBF powder tensile properties in Figure 5.31 were similar to the initial as-received properties generated with the waste L-PBF powder. Average maximum load was only 3.5% higher, while ultimate tensile stress was 0.6% higher, and elastic modulus was 1.4% higher. Although, spatially there was a similar pattern to the used L-PBF powder where the specimens around the edges display a higher strength and lower ductility than those in the central region of the build area. Those edge specimens contributed to the slightly higher averages, where the central specimens generally exhibited equal or lower strength. That is even more evident in the elongation values where those in the center were similar to the initial as-received values, and the values near the edges of the build area pulled the group average down significantly. Unfortunately, the 0.2% offset yield data for most of the specimens was corrupted, and thus was not included in the chart.

A topographical analysis of the tensile fracture surfaces was performed using the Keyence VR-3100 optical profilometer. The resulting surface maps, presented in Figure 5.32, are arranged to reflect the spatial distribution of specimens within the build chamber, thereby facilitating direct comparison with location-dependent mechanical property trends. The fracture surfaces predominantly exhibit the characteristic cup-and-cone morphology associated with ductile failure, evidenced by peripheral elevation or depression of the fracture edges. This geometry arises from localized necking and shear lip formation during tensile loading, indicative of substantial plastic deformation prior to fracture. Although each tensile specimen yields two fracture halves, only one is displayed in the figure due to their symmetrical nature; the opposing half serves merely as a mirrored counterpart and offers no additional topographical insight.

Maximum Load (KN)															
				Group Avg	Initial	% difference									
				63.6	61.5	3.5	Row Avg	% difference from group average value							
64.1	64.4	63.7	63.6	63.8	64.1	64.8	64.1	0.8	1.3	0.3	0.1	0.3	0.8	2.0	
64.1	63.8	No Data	62.8	62.8	63.5	64.3	63.6	0.9	0.3	No Data	-1.3	-1.2	-0.1	1.2	
64.4	63.1	62.5	62.1	62.3	63.0	63.7	63.0	1.3	-0.8	-1.7	-2.4	-2.0	-0.9	0.1	
64.3	62.9	62.5	62.1	62.4	63.3	64.0	63.1	1.2	-1.1	-1.7	-2.2	-1.8	-0.5	0.7	
64.9	64.1	63.9	63.7	63.6	64.1	64.6	64.1	2.2	0.8	0.5	0.2	0.1	0.8	1.7	

Tensile stress at Max Load (MPa)															
				Group Avg	Initial	% difference									
				999	993	0.6	Row Avg	% difference from group average value							
1008	1012	1002	1000	1003	1007	1019	1007	0.8	1.3	0.3	0.1	0.3	0.8	2.0	
1008	1003	No Data	986	987	998	1011	999	0.9	0.3	No Data	-1.3	-1.2	-0.1	1.2	
1012	992	982	975	980	990	1001	990	1.3	-0.8	-1.7	-2.4	-2.0	-0.9	0.1	
1011	989	983	977	981	994	1006	992	1.2	-1.1	-1.7	-2.2	-1.8	-0.5	0.7	
1021	1007	1005	1001	1001	1007	1016	1008	2.2	0.8	0.5	0.2	0.1	0.8	1.7	

Modulus (GPa)															
				Group Avg	Initial	% difference									
				106.2	104.7	1.4	Row Avg	% difference from group average value							
106.2	107.2	108.3	105.3	107.1	113.0	110.7	108.2	0.0	0.9	2.0	-0.9	0.8	6.4	4.3	
107.5	106.0	No Data	100.9	106.7	105.6	110.1	106.1	1.2	-0.2	No Data	-5.0	0.5	-0.5	3.6	
103.3	104.8	104.5	104.1	102.6	106.8	106.1	104.6	-2.7	-1.3	-1.6	-2.0	-3.4	0.6	-0.1	
106.6	103.9	101.2	103.1	105.3	109.4	106.3	105.1	0.4	-2.2	-4.7	-3.0	-0.8	3.1	0.1	
110.1	107.4	105.4	106.3	104.6	104.4	109.9	106.9	3.6	1.1	-0.7	0.1	-1.5	-1.7	3.5	

Elongation (%)															
				Group Avg	Initial	% difference									
				0.134	0.180	-25.3	Row Avg	% difference from group average value							
0.077	0.099	0.107	0.136	0.133	0.152	0.109	0.116	-43.0	-26.2	-20.3	0.9	-0.9	13.3	-19.0	
0.097	0.177	No Data	0.151	0.196	0.152	0.123	0.149	-27.7	31.4	No Data	12.4	45.7	13.0	-8.4	
0.138	0.088	0.151	0.174	0.150	0.152	0.113	0.138	2.4	-34.5	12.0	29.3	11.7	13.3	-15.9	
0.179	0.181	0.179	0.186	0.179	0.147	0.147	0.171	33.1	34.9	33.3	38.6	33.2	9.1	9.3	
0.088	0.093	0.136	0.073	0.109	0.095	0.103	0.100	-34.6	-30.9	1.4	-45.5	-18.8	-29.6	-23.2	

Tensile Yield stress (0.2% Offset) (MPa)															
				Group Avg	Initial	% difference									
				936	913	2.5	Row Avg	% difference from group average value							
935	943					951	943	-0.2	0.7					1.6	
936		No Data				943	940	0.0		No Data				0.8	
941	921					930	931	0.5	-1.6					-0.7	
940	915				929	935	930	0.4	-2.2				-0.8	-0.1	
947	937			930	937	946	939	1.2	0.1			-0.7	0.0	1.1	

Tensile strain at Max Load (mm/mm)															
				Group Avg	Initial	% difference									
				0.063	0.065	-3.3	Row Avg	% difference from group average value							
0.054	0.061	0.063	0.065	0.068	0.061	0.057	0.061	-14.0	-2.7	0.0	2.8	8.0	-2.6	-10.1	
0.055	0.068	No Data	0.071	0.064	0.066	0.058	0.064	-12.7	7.5	No Data	12.4	2.6	5.0	-7.9	
0.063	0.062	0.067	0.070	0.068	0.067	0.060	0.065	0.8	-0.8	6.0	10.6	8.9	6.1	-4.4	
0.059	0.063	0.069	0.068	0.069	0.061	0.061	0.064	-6.1	0.9	9.3	8.2	9.5	-2.9	-3.0	
0.052	0.061	0.064	0.063	0.062	0.061	0.058	0.060	-17.7	-2.6	2.3	0.3	-1.7	-3.6	-8.5	

Figure 5.31. Final virgin EB-PBF Arcam powder tensile properties (colors scales relative to the individual table and labeled “Initial” value).

A pronounced distinction is observed between specimens produced from virgin EB-PBF feedstock and those fabricated using reused L-PBF powder. The EB-PBF powder specimens exhibit greater variability in surface height, showing heterogeneous deformation and energy absorption across the fracture plane, hallmarks of a more ductile material. In contrast, the fracture surfaces of the used L-PBF powder specimens are notably flatter and more uniform, consistent with a more brittle fracture mechanism characterized by limited plasticity and abrupt failure. Additionally, the upper surface diameter of the used L-PBF powder specimens is measurably larger than that of the virgin EB-PBF powder counterparts. This dimensional disparity reflects differences in necking behavior, demonstrated by a more brittle material with rapid crack propagation. Unlike the mechanical property data, which demonstrated sensitivity to spatial positioning within the build chamber, the fracture surface topographies do not exhibit any discernible correlation with specimen location. This hints that while thermal history and beam quality may influence bulk mechanical performance, they exert minimal influence on the localized fracture morphology, implying that powder condition and intrinsic material properties dominate the fracture behavior in this context.

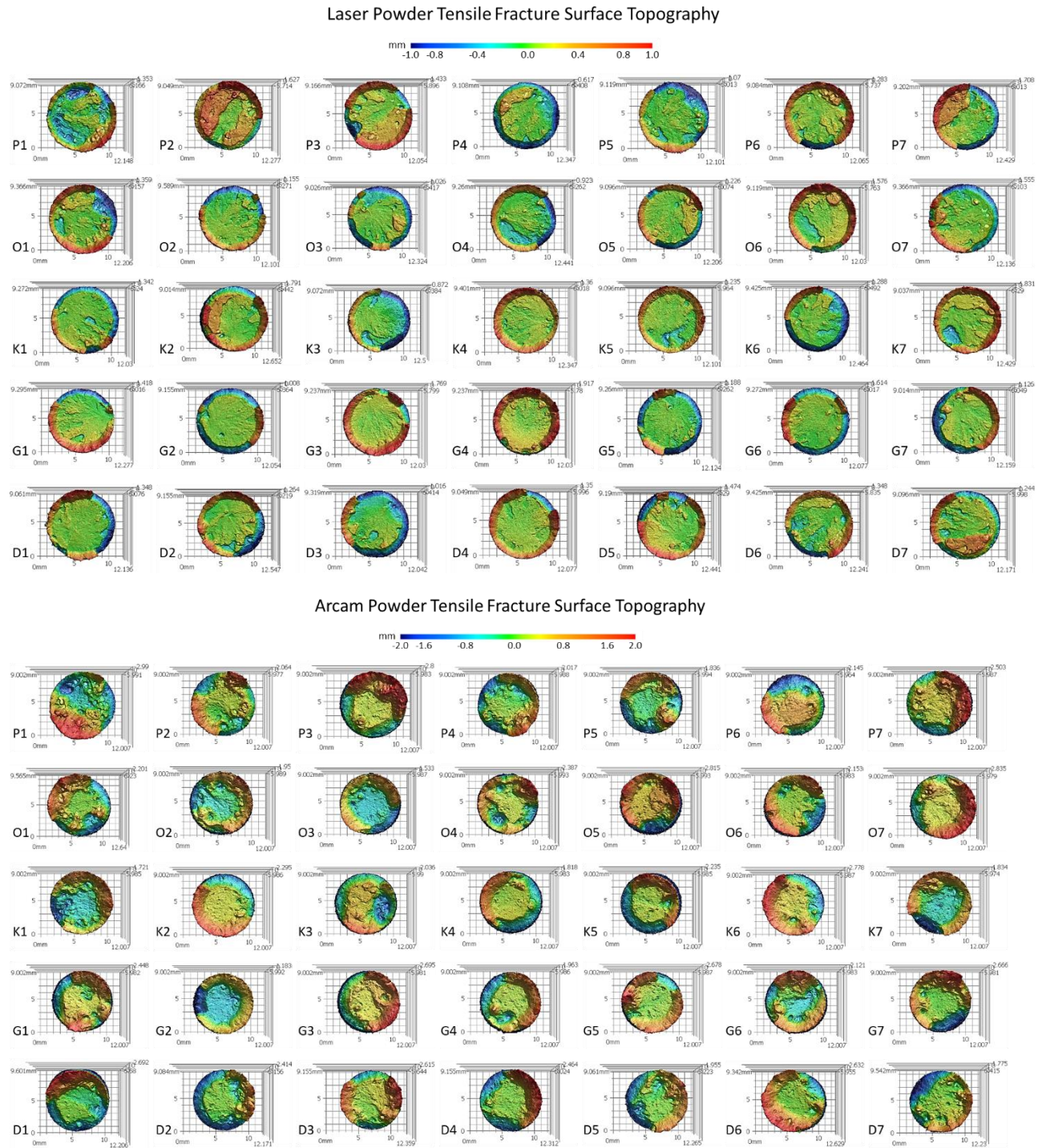


Figure 5.32. Final used L-PBF powder (top), and virgin EB-PBF Arcam powder (bottom), tensile specimen fracture surface topography (one side).

During high-resolution imaging of the fracture surfaces, discrete regions exhibiting pronounced reflectivity were identified across all tensile specimens. These features, which could not be resolved using conventional optical microscopy, are exemplified in Figure 5.33 by a representative sample fabricated from used L-PBF powder (specimen D6). Topographical analysis revealed that these reflective zones are characteristically smooth and planar, typically located either at the apex of surface protrusions or within recessed depressions. In certain instances, such as the area highlighted by the red box, adjacent reflective regions were observed at differing elevations, separated by a steep vertical discontinuity. This abrupt change in surface topology indicates a localized deviation in crack propagation, likely driven by a transition into a structurally weaker region of the material.

The morphology and spatial positioning of these features are consistent with the presence of lack-of-fusion (LOF) defects, which arise from incomplete melting or insufficient bonding between powder layers during the additive manufacturing process. LOF defects are known to act as stress concentrators, promoting premature crack initiation and altering the fracture path. Their smooth, reflective nature is symptomatic of unmolten or partially sintered powder particles that failed to integrate into the surrounding matrix. The occurrence of these defects across all specimens, regardless of powder condition, underscores the sensitivity of the EB-PBF processes to local thermal fluctuations and scan path irregularities. Moreover, the influence of these defects on crack trajectory highlights their potential role in reducing fracture toughness and contributing to anisotropic failure behavior in EB-PBF Ti-6Al-4V components.

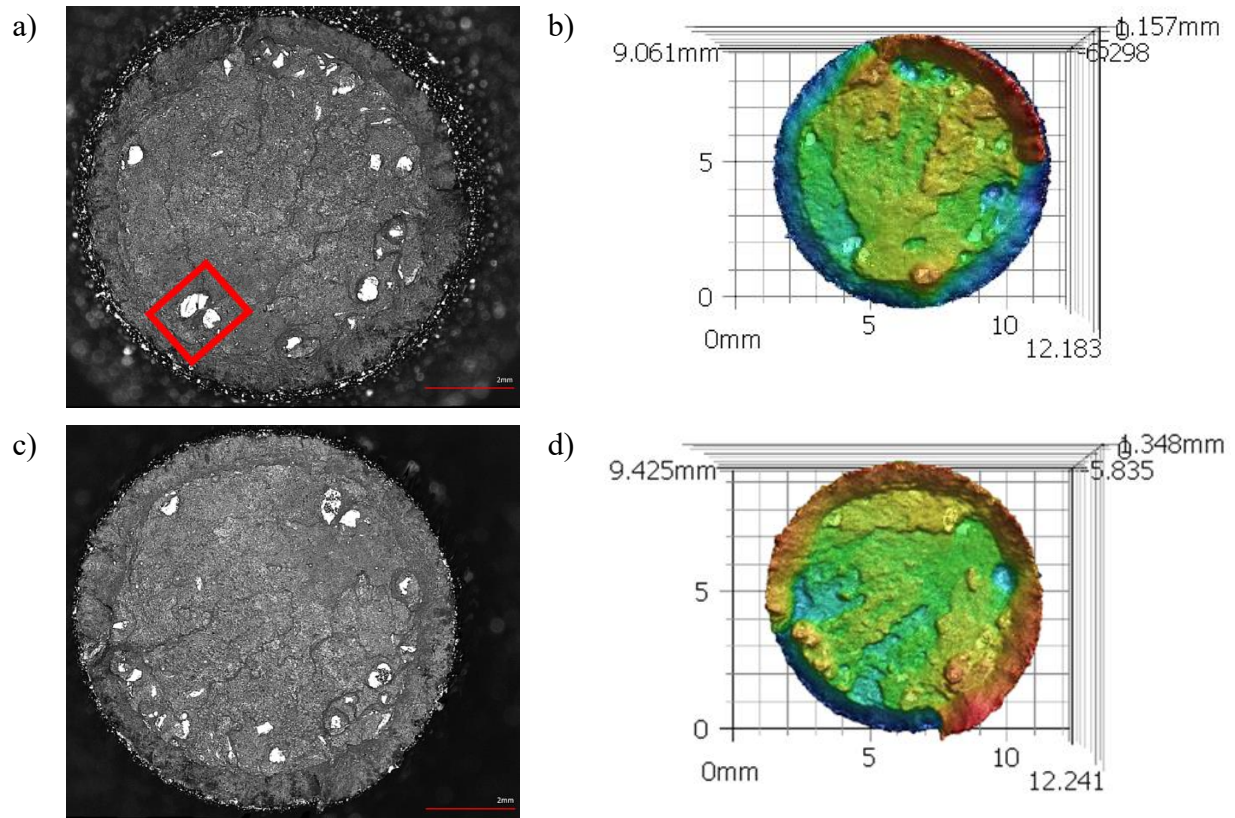


Figure 5.33. Closer look at used L-PBF powder D6 tensile specimen fracture surface, a) side 1 optical image, b) side 1 surface topography, c) side 2 optical image, d) side 2 surface topography. Red box indicates region examined by SEM.

To further characterize the highly reflective regions observed on the fracture surfaces, the D6 sample fabricated from used L-PBF powder was examined using scanning electron microscopy (SEM). Unlike optical microscopy, which relies on photon-based imaging and is limited by surface reflectivity and resolution constraints, SEM utilizes a focused electron beam to generate high-resolution topographical and compositional contrast. This approach effectively mitigates the influence of surface reflectivity, enabling the visualization of sub-micron features that are otherwise obscured. Figure 5.34 presents the SEM micrograph corresponding to the region previously highlighted in Figure 5.33. Under SEM imaging, the rough fracture surface exhibits high brightness, consistent with the presence of microvoid coalescence, a hallmark of ductile

failure mechanisms. In contrast, the previously reflective, optically bright regions now appear dark, indicating a distinct surface morphology and potentially different material response during fracture.

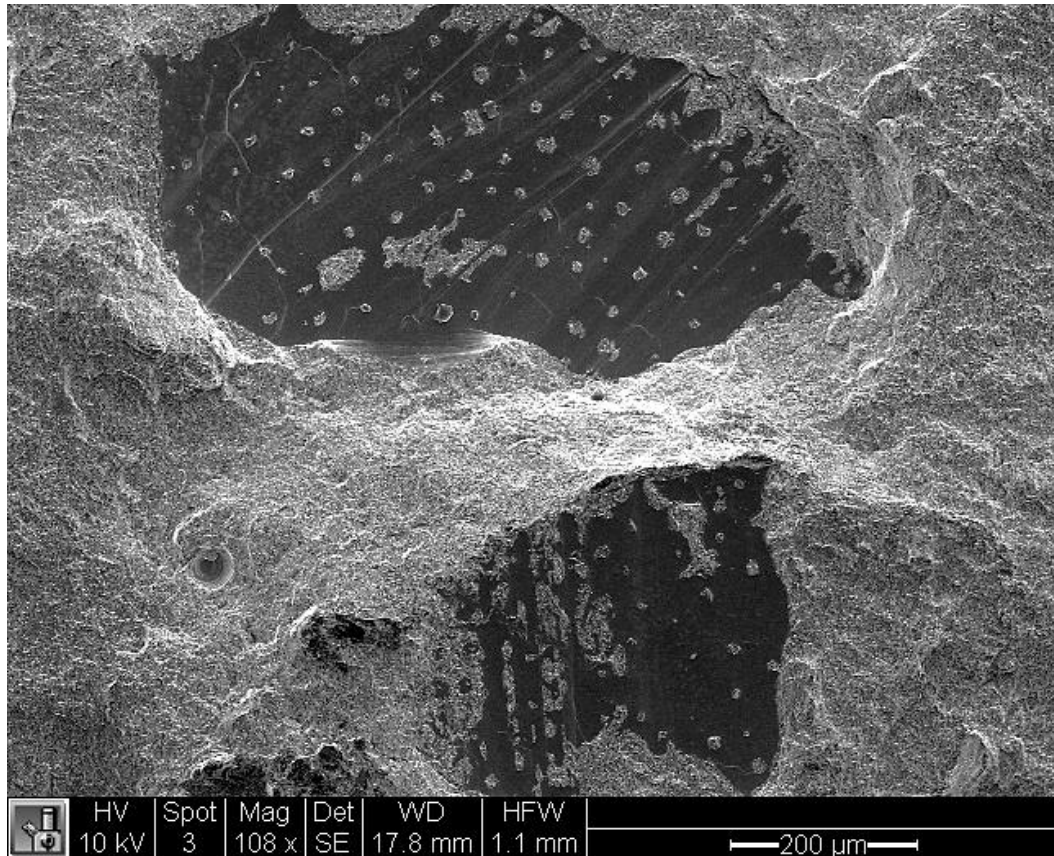


Figure 5.34. SEM image of shiny surfaces in area of used L-PBF powder D6 tensile specimen side 1 fracture surface.

The dark, smooth region resembles a planar basin, analogous to a lake, interspersed with numerous discrete features resembling islands. These "islands" may represent partially fused or unmelted powder particles embedded within the defect zone, further supporting the hypothesis of lack-of-fusion (LOF) defects. Additionally, the presence of linear striations oriented at approximately 45° and 90° across the larger and smaller smooth regions, respectively, displays a directional patterning consistent with the scan strategy of the electron beam during layer

deposition. These lines are likely artifacts of the beam path during the EB-PBF process, reflecting the raster orientation and thermal gradients imposed during solidification. The directional nature of these features may also imply anisotropic bonding conditions at the layer interface, which could exacerbate mechanical discontinuities and facilitate crack propagation along preferred paths. Collectively, these observations reinforce the interpretation that the smooth, reflective regions are manifestations of LOF defects, which compromise interlayer cohesion and serve as preferential sites for fracture initiation and propagation.

Figure 5.35a provides a high-magnification SEM view of one of the previously identified “island” features from the used L-PBF powder specimen D6. It reveals a region of material discontinuity that lacks the characteristic dimple morphology typically associated with ductile fracture in Ti-6Al-4V. Instead, the surface exhibits tearing and smooth planar features, evidence of localized brittle behavior and incomplete fusion. In contrast, Figure 5.35b captures the opposing side of the same fracture surface location from specimen D6. This image reveals unmelted spherical powder particles exhibiting similar surface tearing on their upward-facing sides, further substantiating the hypothesis that the optically reflective regions correspond to LOF defects.

The presence of LOF defects within the fracture surface is consistent with their role as stress concentrators and crack initiation sites. These defects compromise interlayer bonding and introduce mechanical discontinuities that alter the crack path, often redirecting propagation toward structurally weaker zones. The observed planarity of the LOF regions, aligned parallel to the build plane, reinforces expectations based on the layer-wise nature of the PBF process, where insufficient energy input or poor powder packing can result in incomplete melting along horizontal interfaces.

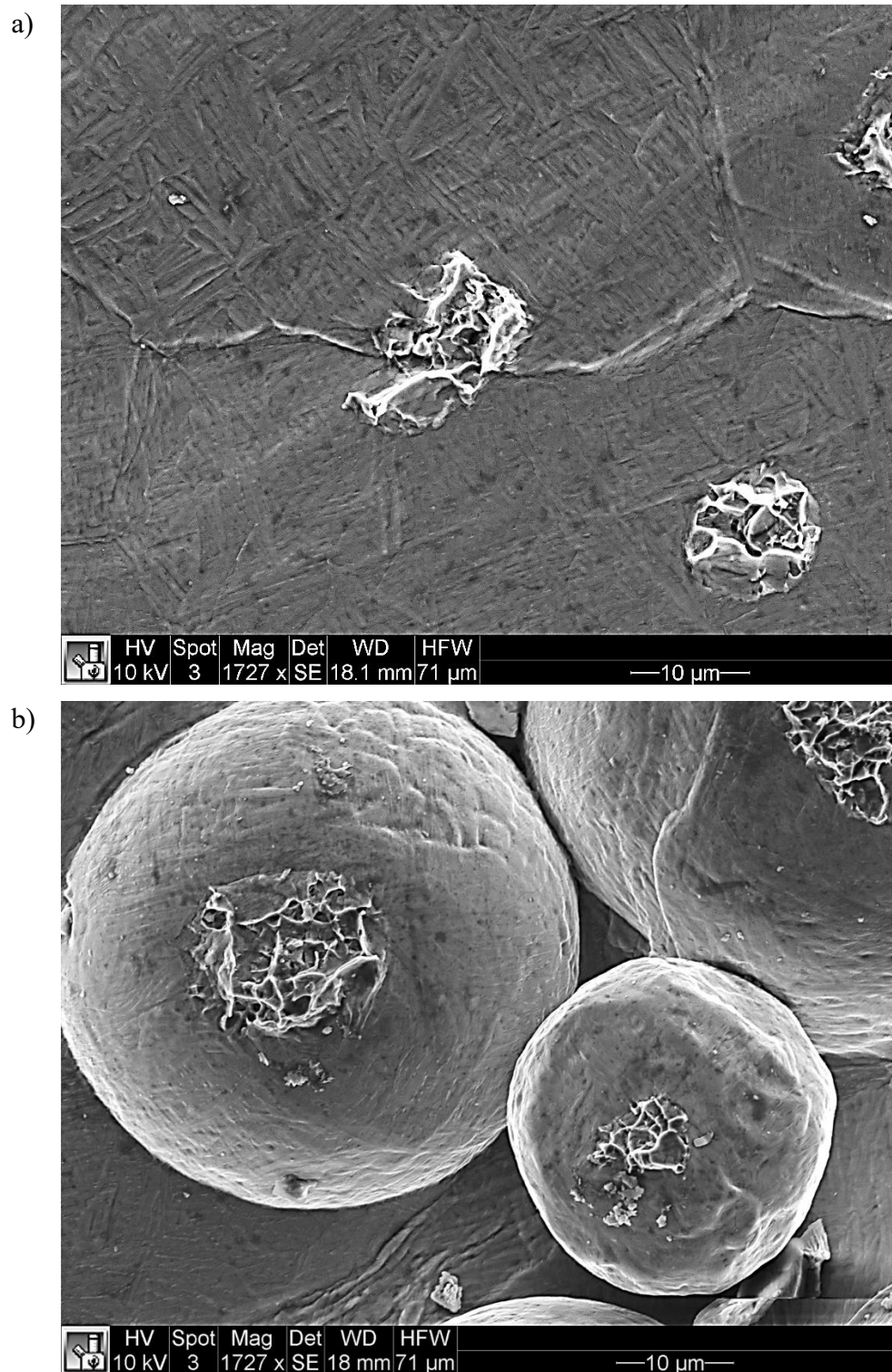


Figure 5.35. SEM images of used L-PBF powder D6 tensile specimen zoomed in on, a) small fracture surface "island" within the flat lack-of-fusion region, and b) opposing side of fracture "island" showing area of partially fused unmelted powder.

An additional noteworthy observation from Figure 5.35a is the visibility of the underlying microstructure without the need for surface polishing or chemical etching. Distinct linear boundaries delineating prior- β grains are clearly discernible, containing the characteristic ($\alpha + \beta$) basketweave morphology typical of Ti-6Al-4V processed via EB-PBF. The grain structure appears equiaxed in nature, which is consistent with the vertical build orientation of the specimen and the imaging perspective aligned with the build direction. In this orientation, columnar grains extend into or out of the viewing plane, resulting in a cross-sectional appearance that resembles equiaxed morphology. This microstructural clarity not only confirms the thermal history and solidification behavior of the material but also highlights the potential for direct microstructural assessment from fracture surfaces in EB-PBF components.

Figure 5.36 presents a detailed microstructural characterization of specimen D6, fabricated from the used L-PBF powder. A segment from the grip region of the specimen was sectioned, mechanically polished, chemically etched, and subsequently examined under SEM at high magnification. The resulting micrograph reveals a well-defined lamellar morphology comprising alternating α and β phases, consistent with the microstructural features typically observed in EB-PBF Ti-6Al-4V alloys and corroborated by existing literature.

The microstructure exhibits a Widmanstätten, basket-weave pattern, characterized by lamellar α phase colonies nucleated within prior- β grains. These prior- β grain boundaries are clearly delineated and are enveloped by the lamellar α phase, indicating a diffusion-controlled solid-state transformation during post-solidification cooling. The presence of this microstructural arrangement is evidence that the thermal gradients and cooling rates inherent to the EB-PBF process were sufficient to promote the formation of α lamellae within the retained β matrix, a hallmark of the alloy's response to rapid solidification followed by slower cooling.

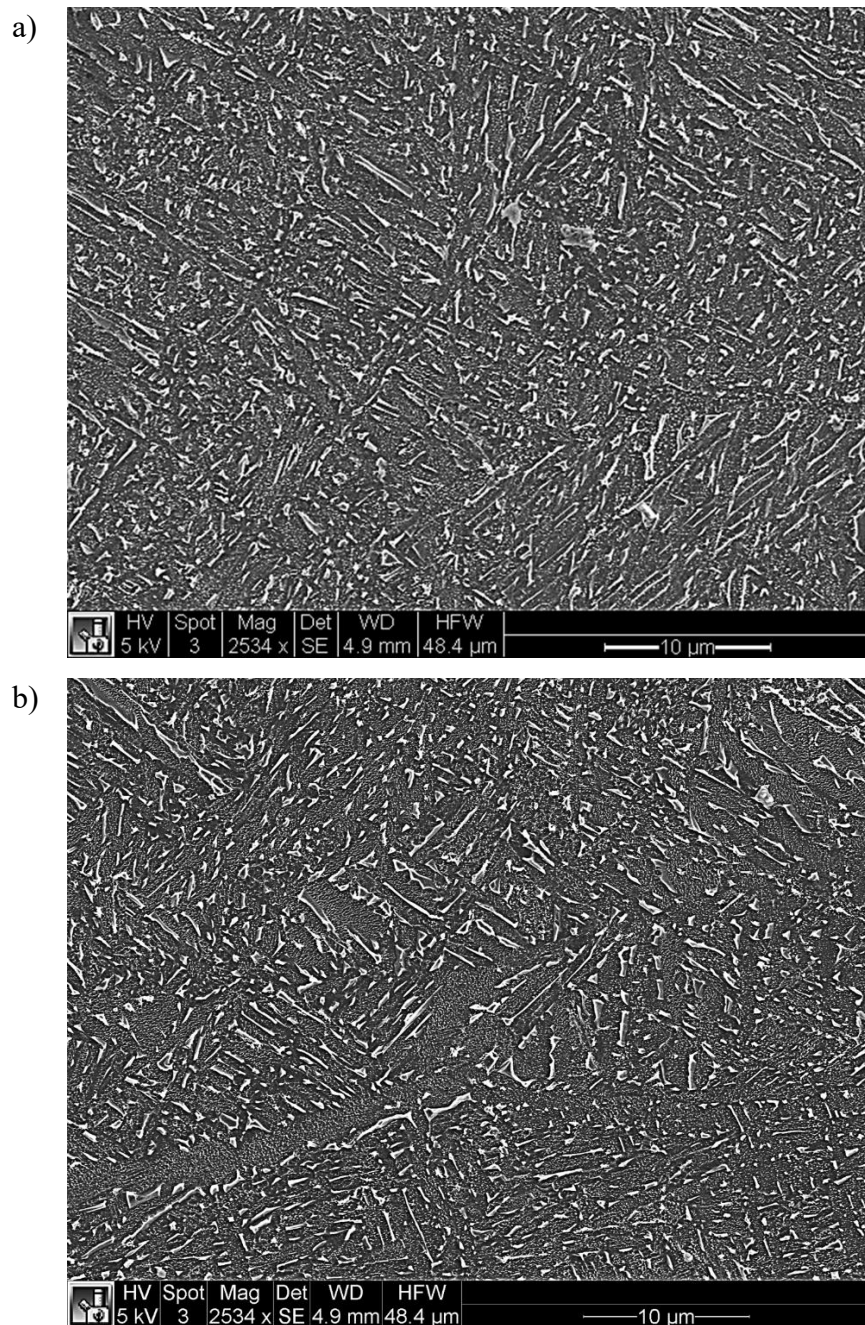


Figure 5.36. SEM images of used L-PBF powder D6 tensile specimen microstructure.

This observation not only confirms the integrity of the microstructure in the used L-PBF powder specimen, but also provides insight into the thermal history and phase evolution during

the build process. The clarity and consistency of the lamellar $\alpha + \beta$ structure reinforce the notion that reused powder, under controlled processing conditions, can still yield microstructural fidelity comparable to virgin powder. Furthermore, the delineation of prior- β grain boundaries offers a valuable reference for assessing grain morphology, orientation, and potential anisotropy in mechanical behavior.

For comparative purposes, a similar topographical analysis was conducted on the P1 tensile specimen fabricated using virgin EB-PBF Arcam powder. As shown in Figure 5.37, the fracture surface exhibits a notably higher density of highly reflective features relative to the specimens produced from used L-PBF powder. The prevalence of these features suggests a more complex and discontinuous fracture path, consistent with the behavior of a ductile material undergoing significant plastic deformation. The increased number of reflective regions implies that the crack propagation mechanism was influenced by the material's ability to absorb and redistribute strain energy. Rather than following a singular, linear trajectory, the fracture path appears to have intermittently deviated toward localized LOF weak points, facilitating continued propagation through successive regions of reduced resistance.

This behavior is characteristic of a ductile fracture, where the material undergoes extensive strain to failure, resulting in a more tortuous and irregular fracture surface. The increased surface topography variation observed in the P1 specimen further supports this interpretation. These results reinforce the correlation between powder condition and fracture behavior, with virgin EB-PBF powder yielding a microstructure that promotes energy dissipation and delayed crack propagation, thereby enhancing ductility at the expense of ultimate strength. The observed size and quantity of process-induced LOF voids that existed within the P1 sample speaks to the lower

beam quality that occurs at the boundaries of the build area, as P1 exhibited one of the lowest strains to failure.

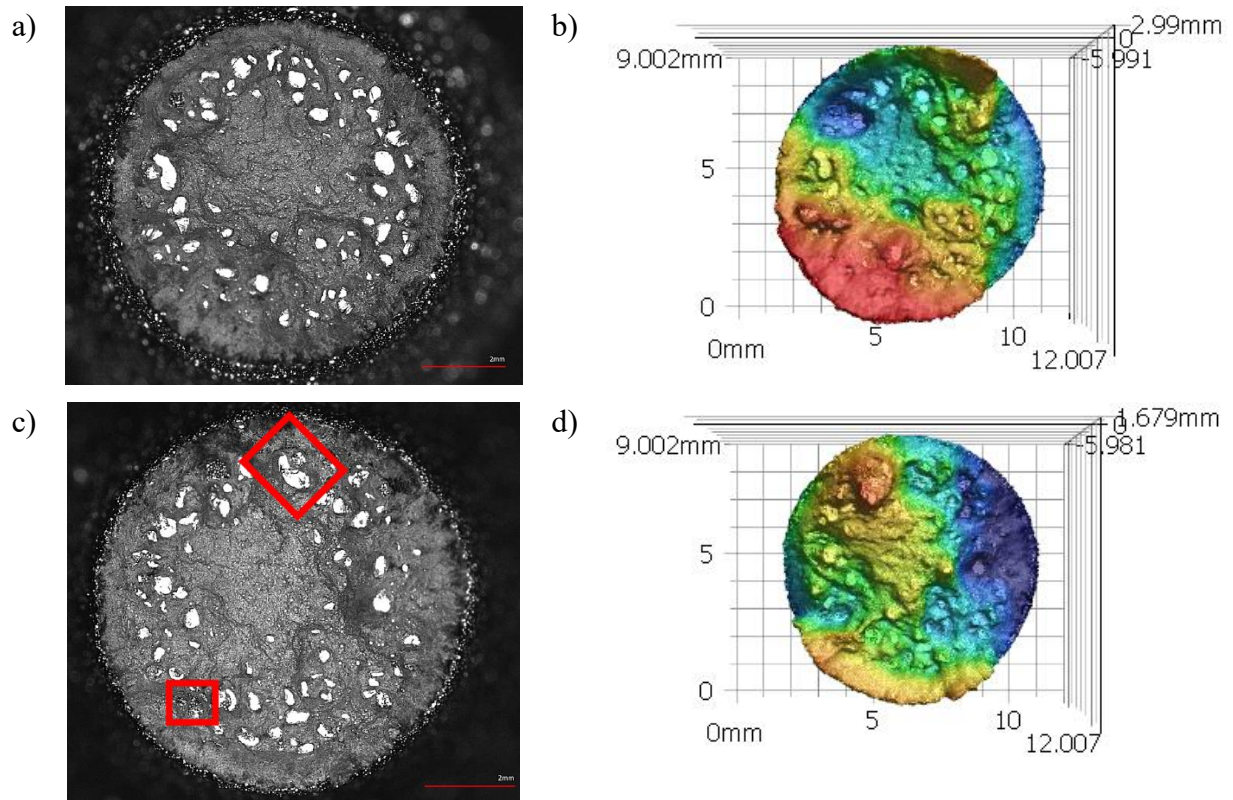


Figure 5.37. Closer look at virgin Arcam P1 tensile specimen fracture surface, a) side 1 optical image, b) side 1 surface topography, c) side 2 optical image, d) side 2 surface topography. Red boxes indicate regions examined by SEM.

The regions highlighted in Figure 5.37, indicated by red boxes, were also subjected to detailed examination using SEM. The larger boxed area was analyzed in Figure 5.38, while the smaller region was investigated in Figure 5.39. SEM imaging of the P1 specimen, fabricated from virgin EB-PBF powder, revealed surface features that closely resemble those previously observed in Figure 5.34 and Figure 5.35 from the used L-PBF powder specimen D6. Surfaces that appeared highly reflective under optical microscopy now present as dark, planar regions under SEM, with topographies reminiscent of smooth basins interspersed with discrete particulate features,

analogous to lakes dotted with islands. This transformation in visual contrast further underscores the limitations of optical imaging in resolving fine-scale morphological details and confirms the presence of localized surface discontinuities. The dark, smooth regions are telling of LOF defects, characterized by insufficient melting and poor interlayer bonding.

Additionally, Figure 5.37b revealed the underlying microstructure of the P1 specimen, where prior- β grain boundaries are clearly visible and exhibit an equiaxed morphology. This observation is consistent with the vertical build orientation; wherein columnar grains extend along the build axis and appear equiaxed when viewed in cross-section. Figure 5.39 further substantiates the presence of LOF defects by displaying a cluster of unmelted powder particles embedded within a large void. These particles retained their spherical morphology, confirm incomplete fusion, and highlight the sensitivity of the EB-PBF process to localized thermal fluctuations and scan path irregularities.

Collectively, these findings reinforce the interpretation that both virgin and used powder feedstocks are susceptible to process induced LOF defects, which manifest as topographical anomalies and contribute to fracture surface complexity. Their presence, regardless of powder condition, emphasizes the importance of process optimization and defect mitigation strategies in achieving consistent mechanical performance in EB-PBF Ti-6Al-4V components. Optical images and topographical analysis of the other tensile specimens not presented in this section can be found in Appendix D.

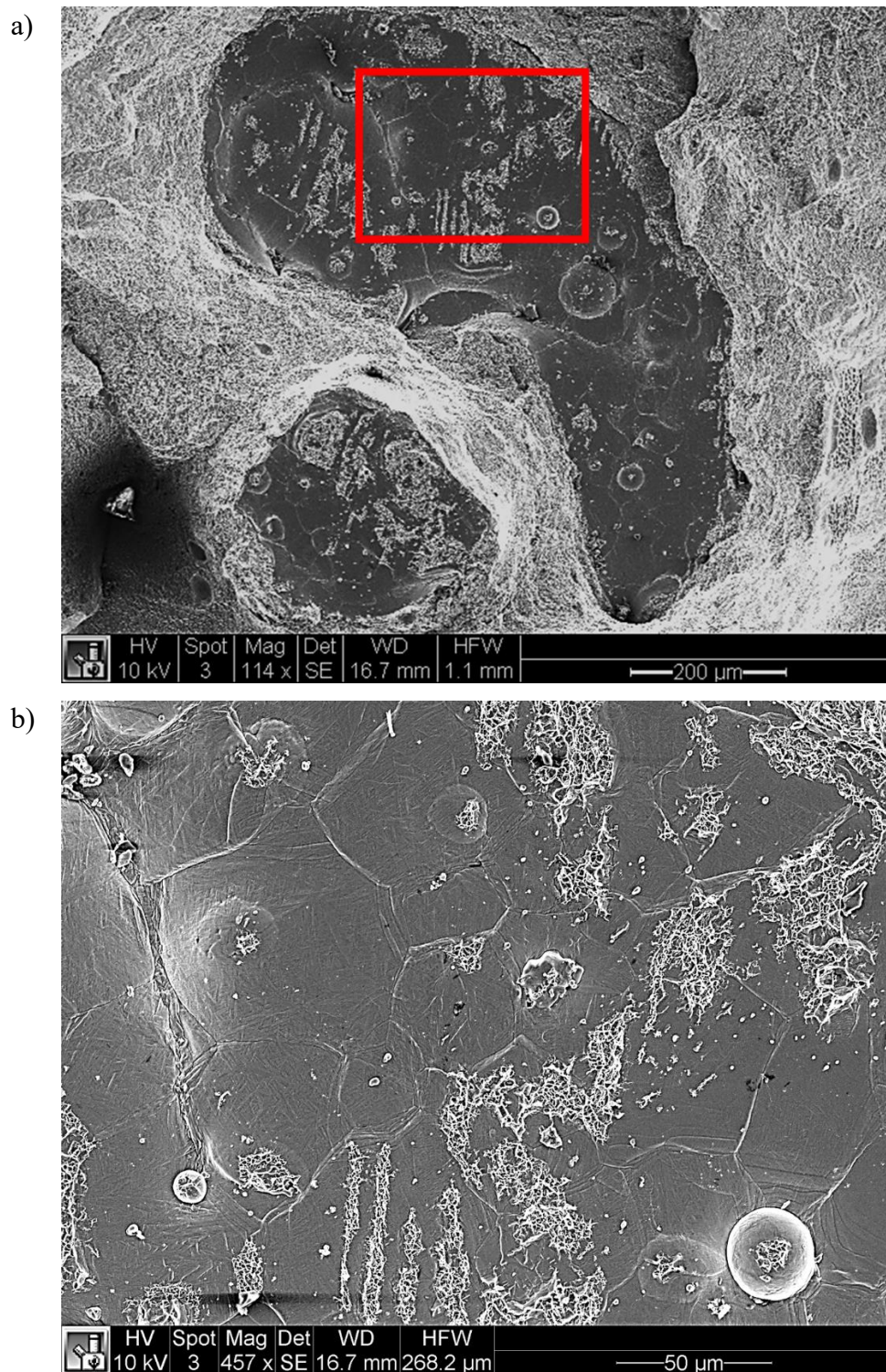


Figure 5.38. SEM images of, a) shiny surfaces in upper area of virgin EB-PBF Arcam powder P1 tensile specimen side 2 fracture surface, and b) high resolution close up of highlighted region from a).

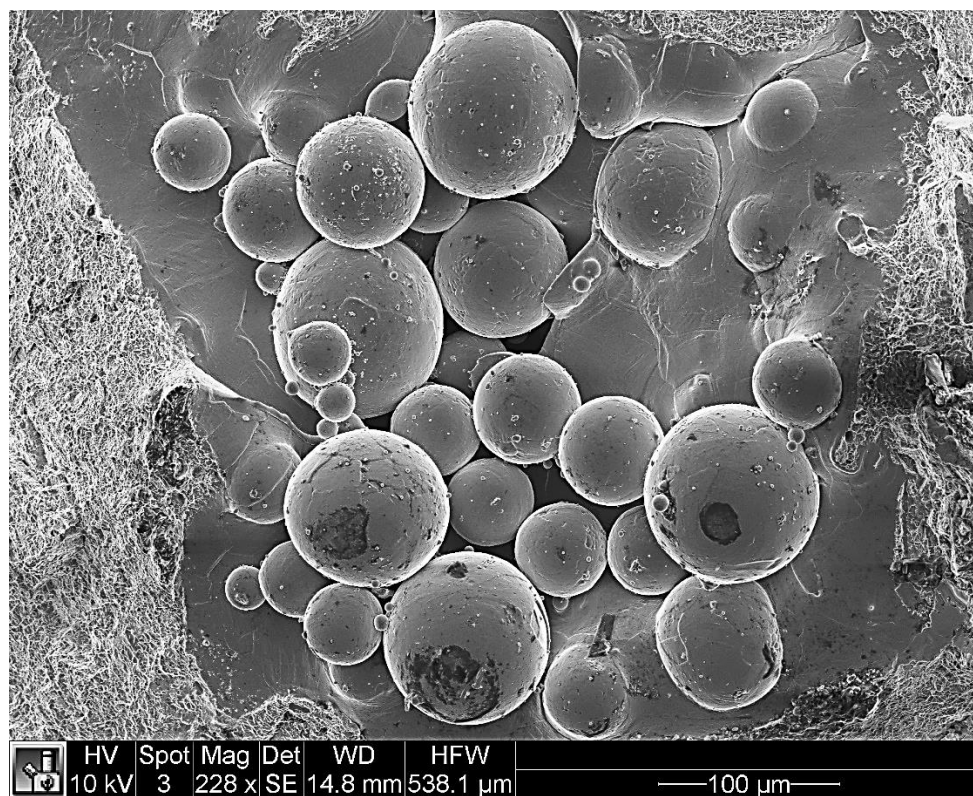


Figure 5.39. SEM image of unmelted powder within a lack-of-fusion defect on the fracture surface of virgin EB-PBF Arcam powder P1 tensile specimen side 2.

Chapter 6. DISCUSSION

This chapter constitutes a pivotal element of this work, wherein the experimental findings are critically examined and contextualized within the framework of existing scientific literature. It facilitates a comprehensive interpretation of the results, emphasizing their relevance to current research and their implications for both theoretical advancement and practical application. The discussion will address the strengths and limitations inherent to the methodologies employed, while underscoring the novel contributions introduced by this investigation. By synthesizing the empirical data with broader disciplinary insights, this chapter aims to construct a cohesive narrative that articulates the significance of the work and its potential impact on the academic community and industrial practice.

6.1 FEEDSTOCK ANALYSIS

A rigorous analysis of the material feedstock is a foundational aspect of process reliability and product integrity across advanced manufacturing domains, including additive manufacturing, chemical engineering, and materials science. In the context of this study, characterizing the chemical composition and physical properties of the Ti-6Al-4V feedstock is essential to ensure conformity with the stringent requirements of EB-PBF fabrication. Such analysis underpins the production of components with consistent microstructural and mechanical performance, while mitigating the risk of defects arising from feedstock variability. Given the uncertain pedigree and prior usage history of the waste L-PBF material, feedstock evaluation was particularly critical for identifying potential contaminants or compositional deviations that may have influenced experimental outcomes.

In addition to its role in quality assurance, feedstock analysis was instrumental in advancing cost-efficiency and sustainability objectives within this research. A comprehensive evaluation of

the material inputs enabled the validation of feedstock repurposing strategies, thereby facilitating waste minimization and enhancing overall manufacturing efficiency. The potential to redirect waste streams from one additive manufacturing modality as viable input for another presents significant economic advantages and environmental benefits. This is particularly pertinent in the case of Ti-6Al-4V powder, which is classified as a hazardous material due to its reactive nature and associated handling, production, and disposal costs. Furthermore, detailed characterization of feedstock properties contributes to the informed development of novel materials and processing routes that align with environmentally responsible and economically feasible manufacturing examples. Accordingly, this analysis served as an initial element in ensuring the integrity, resource optimization, and sustainability of the experimental framework.

Chemical analysis was an essential step in verifying the identity and integrity of the waste L-PBF feedstock, particularly given its uncertain background and the elevated risk of misclassification or contamination due to its prior discarding. The compositional assessment confirmed that the material was indeed Ti-6Al-4V, thereby mitigating concerns regarding alloy misidentification. However, minor deviations from nominal specifications were observed: vanadium content measured at 4.8 wt%, slightly exceeding the standard upper limit of 4.5 wt%, while aluminum content registered at 5.1 wt%, marginally below the minimum threshold of 5.5 wt%. These variations, though modest, warrant consideration in the context of alloy performance, as compositional shifts can influence phase stability and mechanical behavior.

Of particular importance in titanium alloy systems is the potential for oxygen uptake during processing and storage. Oxygen, as an interstitial element, promotes the formation of α -phase and increases strength at the expense of ductility, often leading to embrittlement and compromised structural reliability. Surface oxidation, especially in powder form, can exacerbate this effect due

to the high surface area-to-volume ratio. Fortunately, the chemical analysis indicated that the recycled powder had not undergone significant oxygen absorption, meaning the storage conditions and handling practices were sufficient to preserve its metallurgical quality. This finding provided assurance that the feedstock remained suitable for experimental specimen fabrication and that its use would not introduce unintended variability into the mechanical or microstructural results.

Powder flowability is a critical parameter influencing the quality and consistency of builds in EB-PBF systems. In these machines, powder is stored in hoppers flanking the build chamber and is gravity-fed to a raking mechanism that distributes the material across the build platform. The uniformity of this distribution is governed by interparticle friction and cohesion, which are in turn influenced by particle morphology, particularly sphericity. Irregularly shaped particles tend to impede flow and lead to non-uniform layering, which can compromise part integrity. Consequently, post-processing steps such as sieving and PRS filtration are essential to eliminate fines and morphologically deficient particles.

Given the uncertain pedigree of the waste L-PBF powder, its flow characteristics were initially unknown. In its as-received condition, the powder exhibited a Hall flow rate of 27 seconds per 50 grams, aligning with typical values for L-PBF powders characterized by finer particle size distributions and increased surface area, which generally reduce flowability. Following multiple passes through the PRS and sieving operations, the flow rate improved to 24 seconds, which is consistent with EB-PBF powder standards. This enhancement in flow behavior indicates that morphological refinement through mechanical filtering can effectively repurpose waste L-PBF powder for EB-PBF applications, thereby restoring its suitability for consistent layer deposition and reliable machine performance.

Morphological characterization via particle size distribution (D10/D50/D90) analysis further substantiated the initial observations regarding powder quality. In its as-received condition, the material exhibited a disproportionately high fraction of fine particles, along with the presence of oversized contaminants. The Arcam PRS effectively reduced the fine particle content, while subsequent sieving operations successfully removed most of the larger agglomerates. Despite these interventions, the processed powder retained a PSD range of 20-51 μm , which remains significantly narrower and finer than the standard EB-PBF specification of 45-106 μm .

This deviation in particle size was nontrivial, as it directly influenced powder behavior during recoating and beam exposure. Finer particles exhibit increased surface area and reduced mass, making them more susceptible to electrostatic charging and less responsive to gravitational settling. Consequently, the observed PSD explains the powder's heightened sensitivity to "smoking" phenomena, where electrostatic repulsion and beam interaction caused powder displacement, underscoring the importance of particle size optimization for stable EB-PBF operation.

In summary, the comprehensive feedstock analysis demonstrated that the recycled L-PBF powder, despite its initial uncertainty and deviation from EB-PBF specifications, could be effectively conditioned for reuse through specific processing steps. Chemical composition verification confirmed the material as Ti-6Al-4V, with only minor deviations in alloying elements and no significant oxygen uptake, defining its metallurgical integrity. Morphological refinement via PRS and sieving improved flowability and mitigated risks associated with electrostatic instability, while particle size distribution analysis highlighted the importance of eliminating fines and agglomerates to achieve EB-PBF-compatible behavior. Collectively, these findings validate

the potential of the waste L-PBF powder to be repurposed for experimental fabrication, provided that rigorous conditioning protocols were maintained.

6.2 DESIGN AND BUILD

The design and fabrication of experimental specimens constitutes a foundational element in ensuring the rigor and credibility of both scientific inquiry and engineering development. In the context of structural optimization, particularly for metallic sandwich structures incorporating novel core geometries, specimen fidelity directly influences the validity of experimental outcomes. Precise control over specimen dimensions, material composition, and spatial configuration within the build chamber enables systematic isolation of key variables, thereby facilitating robust hypothesis testing. In metal additive manufacturing, where the synthesis of material and geometry occurs concurrently, dimensional accuracy and process consistency are paramount. Such precision is essential not only for achieving reproducible mechanical performance but also for establishing a reliable empirical basis from which meaningful interpretations can be drawn and substantive advancements in the field can be realized.

Construction of the specimens for the experiments in this work was vital to simulating real-world conditions and applications relevant to aerospace and other industries. In engineering and materials science, specimens must mimic the properties and behaviors of actual components of systems to provide insights into their performance, durability, and safety. This involves not only selecting the appropriate materials and fabrication techniques, but also ensuring that the specimens are representative of the conditions they will encounter in practical use. By doing so, potential issues can be identified, designs optimized, and innovative solutions developed that are both effective and efficient. Ultimately, the meticulous design and build of the sandwich specimens for

the 4-point bend experiments underpin the integrity and applicability of the findings in this research, driving progress and innovation across various engineering disciplines.

The advent of PBF additive manufacturing has revolutionized the design and fabrication of advanced metallic sandwich structures, particularly those utilizing Ti-6Al-4V. This technology enables the production of geometrically complex, functionally integrated components wherein the core architecture and face sheets are fabricated as a single monolithic structure, eliminating the need for secondary bonding or joining processes. Such integration offers substantial advantages in terms of structural efficiency, weight reduction, and mechanical reliability, making it highly attractive for aerospace, automotive, and defense applications.

In this study, five distinct lattice topologies were embedded within sandwich panel beams to systematically evaluate their flexural performance and failure mechanisms under bending loads. The selection of these core geometries was informed by their potential to exploit the design freedom inherent to additive manufacturing, particularly in tailoring mechanical response through spatial grading. To optimize the core thickness distribution for a 4-point bend configuration, two computational grading strategies were employed. The first utilized a compliance minimization approach based on the Solid Isotropic Material with Penalty (SIMP) method, which iteratively adjusts material density to enhance stiffness. The second strategy applied a linear-static finite element analysis focused on minimizing peak von Mises stress, thereby targeting improved load distribution and delayed onset of failure. Together, these methodologies facilitated the development of structurally optimized sandwich panels that leverage the full capabilities of PBF technology, while providing a robust foundation for evaluating the interplay between topology, material distribution, and mechanical performance.

The structural design of the sandwich beam specimens was principally informed by the standardized guidelines outlined in ASTM C393, which defines the evaluation of flexural properties in flat sandwich constructions. To accommodate the geometric complexity inherent in architected lattice cores, CAD modeling was conducted using an implicit modeling platform, specifically nTopology. This software proved instrumental in enabling the parametric generation of intricate core geometries, facilitating simulation of 4-point bend load conditions, and integrating the resulting mechanical response data into spatially graded design functions. Through this workflow, a continuously variable lattice thickness distribution was applied across the core domain, thereby leveraging the full potential of additive manufacturing for performance optimization.

Although the face sheet thickness was maintained as a constant parameter across all test specimens to isolate the influence of core architecture, the modeling framework readily supports the implementation of spatially varying face sheet thicknesses. Such an extension could yield further enhancements in mechanical efficiency, particularly under constraints of constant relative density, by enabling localized reinforcement in regions of elevated stress concentration.

Moreover, the spatial relationship between core geometry and loading points warrants careful consideration in future design iterations. For instance, the Primitive lattice topology exhibited suboptimal performance due to the orientation of unsupported face sheet regions perpendicular to the primary compressive stress vectors, resulting in premature local buckling. Minor modifications to the core layout, such as rotational alignment or strategic reinforcement, could mitigate these effects. Conversely, the Octet lattice configuration appeared to benefit from favorable alignment with the test fixture, potentially skewing comparative performance metrics.

A more neutral and standardized alignment strategy across all specimens would be advisable to ensure equitable evaluation of core topologies under identical loading conditions.

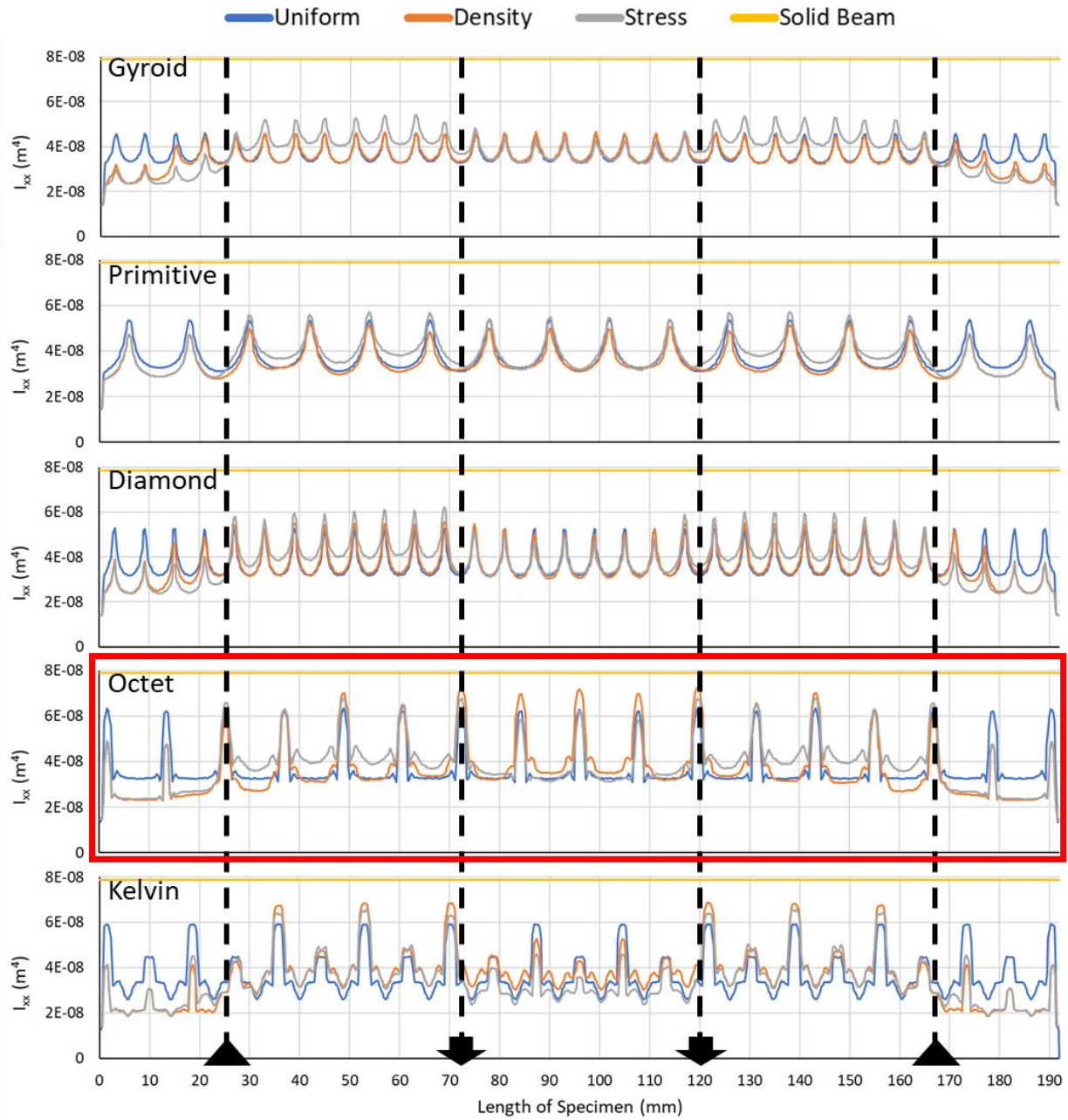


Figure 6.1. Second area moment of inertia I_{xx} analysis for lattice core sandwich beams highlighting the Octet in relation to the load and support locations.

When the moment of inertia was plotted along the beam lengths in relation to the load and support positions of the four-point bending setup (Figure 6.1), a clear pattern emerged: the Octet lattice exhibited peaks of maximum stiffness that closely coincided with the location of the test apparatus. Due to the non-uniform cross-sectional geometry of the beam specimens, all plots revealed a series of repeating peaks and valleys. However, the Octet structure stood out with particularly sharp peaks and abrupt declines, revealing a heightened sensitivity to positional alignment. This means that even minor misalignments during testing could lead to significant variability in mechanical performance. Furthermore, the moment of inertia analysis revealed that within the region between the loading and support points, stress grading substantially increased the minimum bending stiffness compared to both uniform and density-graded configurations. This reinforces the analytical evidence that stress-based grading enhances structural efficiency by elevating baseline stiffness and reducing susceptibility to localized deformation.

The fabrication of sandwich beam specimens via EB-PBF using Ti-6Al-4V powder necessitated a series of iterative trials to establish optimal process parameters. A critical aspect of this effort involved determining the most favorable build orientation within the chamber to ensure consistent material quality and geometric fidelity. Additionally, the selection of an appropriate lattice unit cell size was essential to facilitate complete evacuation of sintered powder from the internal core regions, thereby ensuring consistent structural performance.

Specimens were designed and manufactured at their nominal dimensions; however, the inherent thermal contraction associated with the coefficient of thermal expansion (CTE) of Ti-6Al-4V during cooling led to predictable dimensional shrinkage. This phenomenon was accounted for in post-processing and measurement protocols. Notably, horizontal orientation of the parts within the build chamber was found to significantly mitigate the risk of build failure. This was

particularly evident during the second build iteration, wherein a software malfunction interrupted the completion of the top layers of stress-graded specimens. Due to their horizontal layout, the uniform and density-graded specimens remained intact and usable for mechanical testing. Had the components been oriented vertically, the entire build would have been compromised, resulting in the loss of all specimens.

Surface roughness of the as-built components was observed to be relatively high, consistent with existing literature on EB-PBF processing of Ti-6Al-4V. This characteristic surface morphology was one of the motivating factors for employing macro-scale lattice architectures, wherein the mechanical behavior is governed predominantly by geometric configuration and bulk material properties, rather than surface-driven phenomena. By designing structures at this scale, the influence of surface irregularities on mechanical performance was effectively minimized, allowing for a more accurate assessment of core topology and grading strategies.

The design and build process for the Ti-6Al-4V sandwich beam specimens highlighted the critical interplay between computational modeling, additive manufacturing constraints, and mechanical performance objectives. By leveraging implicit modeling tools and adhering to ASTM C393 standards, this research was able to generate complex lattice architectures with functional grading tailored to 4-point bend conditions. Strategic decisions regarding build orientation and unit cell sizing were essential to ensuring manufacturability and powder removal, while also mitigating risks associated with thermal distortion and build failure. Despite challenges such as surface roughness and software interruptions, the fabrication approach successfully demonstrated the feasibility of producing structurally optimized, monolithic sandwich panels. These outcomes underscore the importance of integrating design flexibility with process-aware strategies to fully exploit the capabilities of EB-PBF in developing next-generation lightweight metallic structures.

6.3 FLEXURAL PROPERTIES

The flexural behavior of the additively manufactured Ti-6Al-4V sandwich beam specimens provides critical insight into the mechanical performance of architected lattice cores under bending loads. This section examines the influence of core topology, spatial grading strategies, and build orientation on the flexural stiffness, strength, and failure mechanisms observed during 4-point bend experiments. By comparing uniform, density-graded, and stress-graded configurations across five distinct lattice geometries, the discussion aims to expose the structural advantages and limitations inherent to each design. Particular attention is given to deformation patterns, load-displacement responses, and the role of localized stress concentrations, offering a comprehensive evaluation of how additive design freedoms translate into functional mechanical outcomes.

Figure 6.2 presents a comparative analysis of the average flexural performance metrics for the fifteen distinct sandwich beam configurations, encompassing five core topologies across three grading strategies. Among all specimens, the Diamond lattice exhibited the highest maximum load capacity, followed closely by the Gyroid structure. Notably, even the uniform Gyroid core surpassed the performance of most other graded configurations, with the exception of the stress-graded Octet, which demonstrated competitive load-bearing capability. The Primitive lattice consistently underperformed, as anticipated from its geometric limitations, while the Octet topology displayed the greatest variability in mechanical response, likely due to the sensitivity to column buckling relative to alignment with the test apparatus.

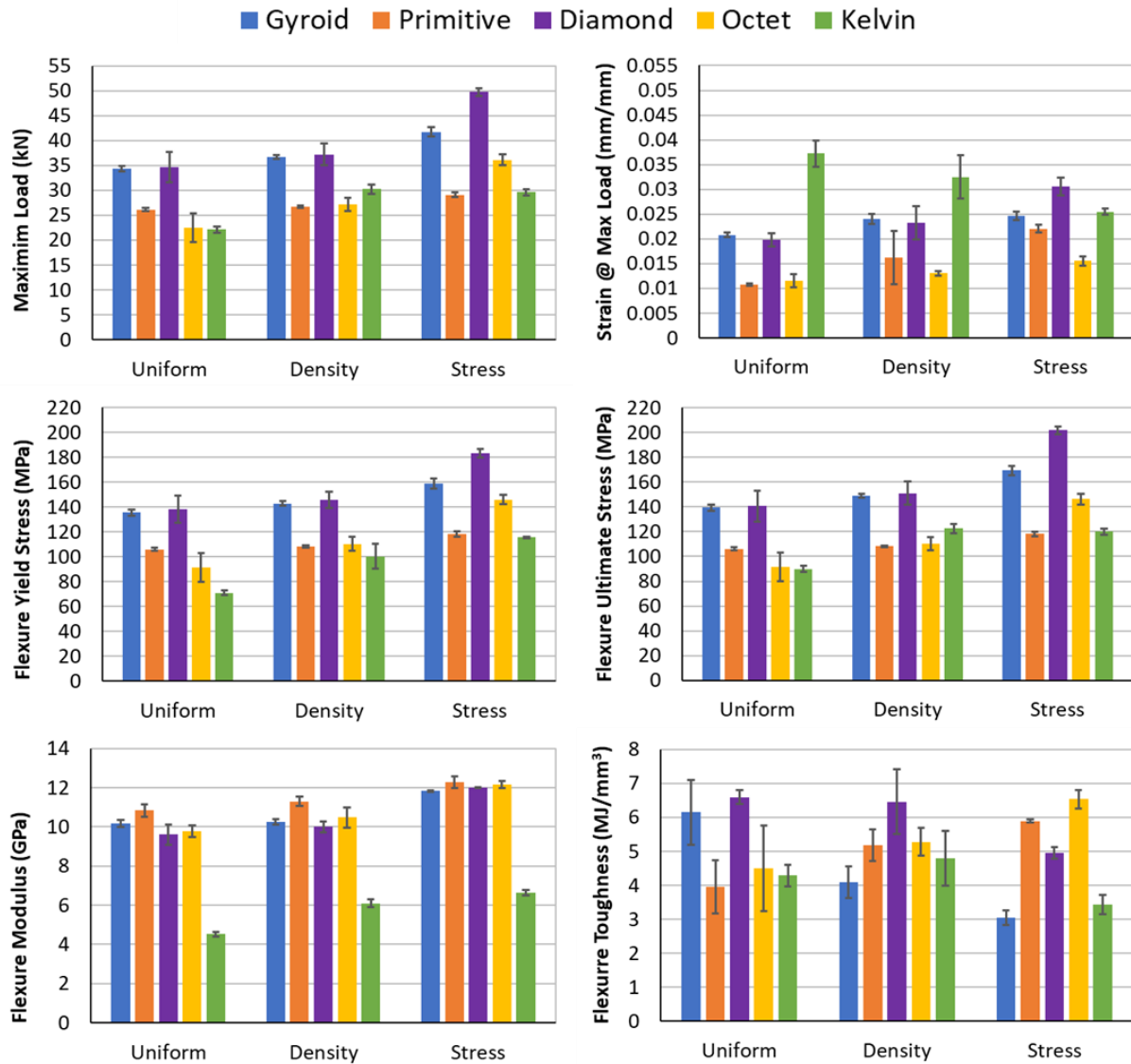


Figure 6.2. Comparison of key flexural properties for the horizontally built Gyroid, Primitive, Diamond, Octet, and Kelvin lattice sandwich structures with the Uniform, Density, and Stress core grading methods, fabricated from waste L-PBF Ti-6Al-4V.

Across the grading methodologies, density grading consistently enhanced maximum load capacity relative to uniform cores, and stress grading yielded further improvements in several cases. However, for the Kelvin topology, the performance gains from stress grading were marginal, with results closely mirroring those of the density-graded variant. When evaluating

strain at maximum load, a markedly different trend emerged: the Kelvin structures exhibited superior strain tolerance, sustaining higher deformation levels prior to softening and failure. This behavior alludes to a more ductile response under flexural loading, potentially attributable to the geometric nature of the Kelvin unit cell.

Furthermore, both density and stress grading approaches contributed to a reduction in the variability observed among uniform cores. Specifically, the flexural strain at maximum load for all stress-graded specimens converged within a narrow range of approximately 0.01 mm/mm, indicating a homogenizing effect of spatial grading on deformation behavior. Analysis of flexural yield stress and ultimate stress revealed similar patterns, with minimal divergence between the two metrics in most cases. Thus, the onset of yielding and peak load capacity occurred in close succession, implying limited post-yield strain softening and a predominantly brittle failure mode across the tested configurations.

The flexural modulus serves as a key indicator of the bending stiffness of sandwich structures, reflecting their ability to resist deformation under applied flexural loads. Across all grading categories, uniform, density-graded, and stress-graded, the TPMS Primitive core consistently exhibited the highest flexural modulus, alluding to superior stiffness relative to the other topologies. The Octet structure, particularly in its graded configurations, also demonstrated elevated stiffness values compared to the Gyroid and Diamond cores. However, it is important to note that the Octet's apparent performance advantage may have been partially influenced by its geometric alignment with the 4-point bend mounting points, which could have introduced favorable load path conditions.

Despite these distinctions, the Gyroid, Primitive, and Octet cores exhibited comparable flexural modulus values, displaying similar resistance to bending across these topologies. In

contrast, the Kelvin structure displayed significantly lower stiffness, approximately half that of the other designs, highlighting its inherently more compliant behavior. This reduction in flexural modulus can be attributed to the Kelvin lattice's nodal arrangement, which places structural nodes away from the primary load paths between the upper and lower face sheets. As a result, stress is distributed through more circuitous routes, leading to increased stress concentrations and reduced load transfer efficiency, as corroborated by finite element analysis (FEA) results.

Regarding the influence of grading strategies, density grading yielded only modest improvements in flexural modulus for the Gyroid, Primitive, Diamond, and Octet cores, implying limited enhancement in stiffness through mass redistribution alone. In contrast, stress grading produced a more pronounced effect, with flexural modulus values increasing by approximately 20% relative to their uniform counterparts. This improvement underscores the effectiveness of stress-informed material distribution in reinforcing critical load-bearing regions, thereby enhancing structural rigidity without altering overall mass or topology.

Flexure toughness T_f is a measure of energy absorption, and it was determined by Equation (6.1) by calculating the area under the flexural stress-strain curve up to the strain at failure ε_r .

$$T_f = \int_0^{\varepsilon_r} \sigma_f d(\varepsilon) \quad (6.1)$$

In general, the implementation of density and stress grading methodologies was observed to reduce the magnitude of uncertainty bands associated with flexural toughness measurements across most lattice core configurations, with the notable exception of the density-graded Kelvin structure. This exception indicates that the Kelvin topology may exhibit greater sensitivity to non-uniform material distribution, potentially due to its inherently less efficient load transfer pathways.

Among the uniform core designs, the Gyroid and Diamond structures exhibited the highest toughness values, demonstrating superior energy absorption capacity under flexural loading.

However, the introduction of grading, particularly stress-based grading, resulted in a marked reduction in toughness for these topologies, thereby diminishing their strain tolerance and overall damage resistance. This outcome highlights a trade-off between stiffness optimization and toughness retention in certain TPMS geometries.

Conversely, the Primitive and Octet lattice cores demonstrated significant improvements in flexural toughness when subjected to grading strategies. Specifically, the graded Primitive core exhibited a 25% increase, while the Octet core achieved a 50% enhancement relative to their respective uniform counterparts. Notably, the stress-graded Primitive core maintained its structural integrity under progressive strain, even after the upper face sheet buckled and lost mechanical support. This sustained load-bearing capacity implies that the measured toughness values may underestimate the true performance potential of the Primitive core, as the experimental test was prematurely terminated before reaching the strain limits observed in the Octet configuration.

The application of grading strategies to the Gyroid and Diamond lattice cores significantly influenced both their elastic and plastic mechanical responses. Specifically, modifications in material distribution through density and stress-based grading altered the internal stress propagation and deformation characteristics under flexural loading. Linear-static finite element model simulations were employed to evaluate the elastic behavior of these graded configurations. The results indicated that stress grading yielded superior performance compared to both density grading and uniform material distribution, while density grading offered only marginal improvements over the baseline uniform design.

Quantitatively, the predicted flexural modulus for the various lattice core geometries, under uniform, density-graded, and stress-graded conditions, all ranged from 113.2-117.3 GPa, as

displayed in Table 6.15. These values are similar to the elastic modulus of the raw material Ti-6Al-4V alloy, and they exhibit no trends relative to the core topology or grading method. Due to the idealization and nature of the linear static analysis, the finite element solver found no difference between the response of the sandwich structures as it would a solid beam. The differences in the values comes from how they were extracted from the mesh, with not all the same nodes in the same locations. This highlights the limitations of a linear-static analysis to provide meaningful predictive capability of lattice sandwich structure behavior.

Table 6.15. Flexural modulus comparison between the linear-static finite element analysis (FEA) predicted values and average experimental values from Figure 6.2

Modulus (GPa)	Uniform		Density		Stress	
	FEA	Exp. (avg)	FEA	Exp. (avg)	FEA	Exp. (avg)
Gyroid	115.9	10.2	113.3	10.3	115.3	11.8
Primitive	115.5	10.8	116.1	11.3	113.2	12.3
Diamond	113.5	9.6	113.6	10.0	115.1	12.0
Octet	113.6	9.8	113.4	10.5	117.3	12.2
Kelvin	114.9	4.5	115.9	6.1	115.7	6.6

The experimental evaluation of the flexural performance of additively manufactured lattice cores revealed more pronounced distinctions among the grading strategies than those predicted by computational modeling. Specifically, stress-strain data obtained from physical testing of horizontally fabricated Diamond specimens demonstrated average flexural modulus of 9.6 GPa, 10.0 GPa, and 12.0 GPa for the uniform, density-graded, and stress-graded configurations, respectively. These empirical values are approximately an order of magnitude lower than those predicted by the linear-static FEA, further underscoring the limitations of idealized numerical models in capturing the nuanced mechanical behavior of real-world structures, as seen in Figure 6.3.

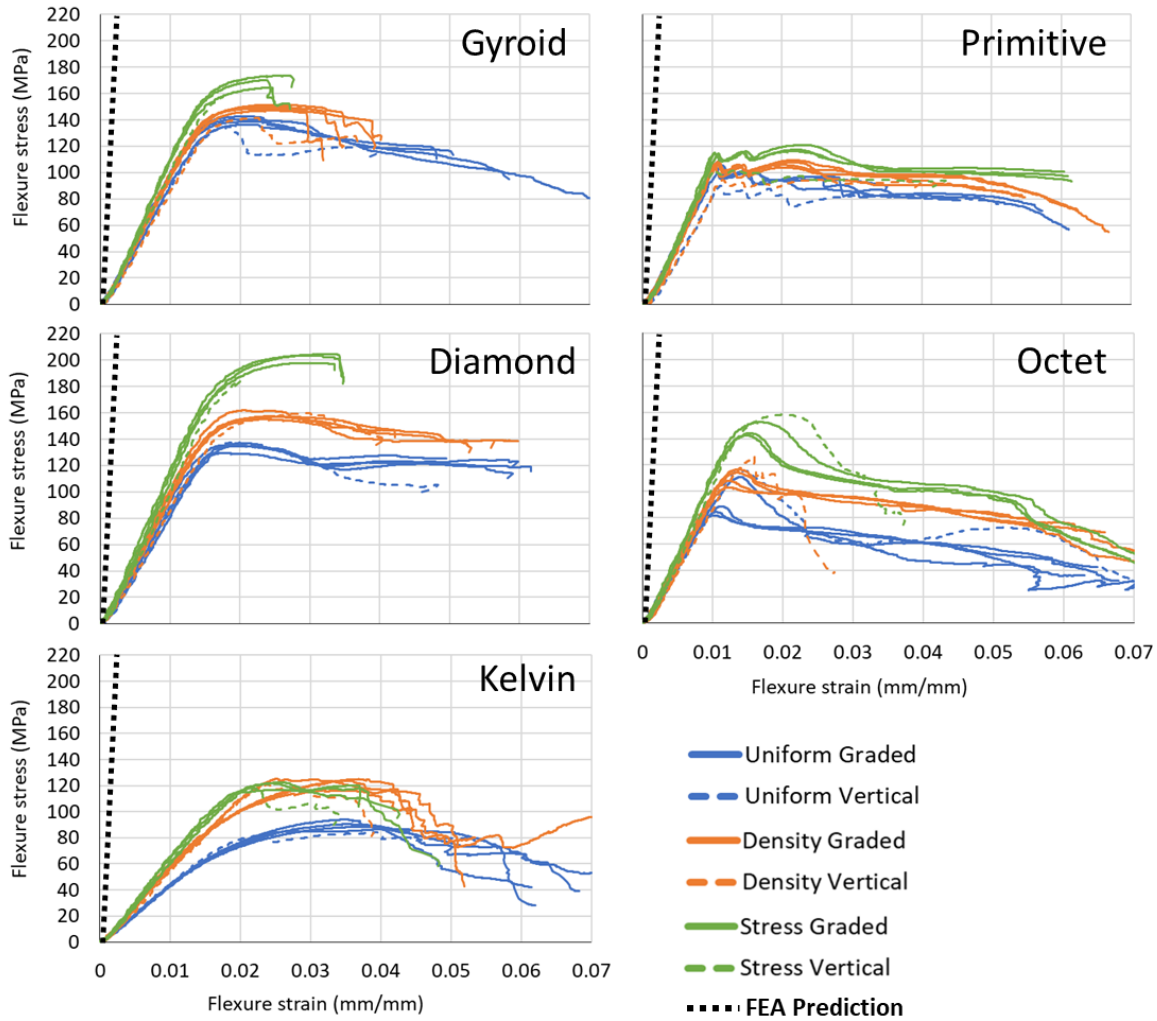


Figure 6.3. Stress-strain results of the 4-point bend experiments with the Finite Element Analysis (FEA) predicted linear-elastic responses.

It is important to distinguish between the flexural modulus and the elastic modulus, or Young's modulus, as the two are often confused. They describe different mechanical behaviors which are derived from different types of tests. Young's modulus is a measure of the material's stiffness under uniaxial tension or compression, and it is a result of an experiment that reflects how much a material resists deformation under those loading conditions. Meanwhile, the flexural modulus is a measure of the material, or material system, stiffness under flexural loads, and it

reflects how much the structure resists bending. This property is influenced by both material properties and geometry, such as moment of inertia, and cross-sectional shape. In practice, it is expected that the flexural modulus and elastic (Young's) modulus would be different, particularly in the context of composite materials, foams, and in this case lattice structures. Bending introduces complex tensile and compressive stress interactions, and geometric effects are not captured in a simple tensile experiment. For example, the flexural modulus is approximately 10% of the Ti-6AL-4V elastic modulus, which also happens to be the relative density of the lattice sandwich structures relative to a solid titanium beam.

A similar explanation comes when considering flexural yield strength of a sandwich structure versus tensile yield strength of a raw material. The difference arises from the fundamentally different stress states, geometry, and failure mechanisms involved in each type of loading. In a tensile experiment, the entire cross section experiences the same stress, while in a bending experiment the stress varies through the thickness of the beam, where for a uniform cross section, the maximum stresses are at the out edges of the face sheets, and zero at the neutral axis. In this way, the face sheets carry most of the tensile and compressive stresses, while the core stabilizes the geometry and resists shear. In tensile experiments, failure is typically uniform yielding across the gauge section, whereas in bending failure can involve buckling, core shear, or face sheet rupture.

In the case of the Diamond configuration, where the lower face sheet failed in tension, it is anticipated that this tensile load was lower than the bulk material yield strength. In a thin sheet-like structure, geometry also plays a significant role in which the failure mechanism can be influenced by out-of-plane deformation, edge effects, localized necking, and surface roughness.

Even from a linear-static FEA, the core geometry created stress concentrations and a non-uniform stress distribution across the width, leading to a lower apparent tensile strength.

Despite the absolute discrepancy, the relative performance trends remained consistent with simulation results. Figure 5.15 presented plots correlating applied load with the percentage of nodes exceeding the yield criterion, offering insight into the structural resistance of various lattice configurations. Among the uniformly graded specimens, the Primitive lattice exhibited the highest resistance, with the Gyroid, Diamond, and Octet, showing comparable performance, and the Kelvin structure demonstrating significantly lower strength. This qualitative trend aligns with the experimental results summarized in Table 6.15, which extend to both density-graded and stress-graded cases. Although the values differ significantly between simulation and experiment, the relative ranking from strongest to weakest remains consistent across both datasets, reinforcing the validity of the observed bending resistance hierarchy.

The divergence between FEA predictions and experimental outcomes can be attributed to several modeling simplifications implemented for computational tractability. These include the use of idealized boundary conditions, reduced mesh density, and the assumption of isotropic, linear-elastic material behavior without incorporating damage evolution or failure criteria. Furthermore, the model did not account for several real-world factors that contribute to compliance and energy dissipation, such as surface roughness at roller-pin contact interfaces, mechanical compliance within the test apparatus, localized buckling of the upper face sheet between load application points, and microstructural damage accumulation within the lattice core. Collectively, these phenomena introduce additional deformation mechanisms that reduce the effective flexural stiffness observed during physical testing.

Figure 6.4 presents a comparative assessment of flexural yield stress, flexural modulus, and toughness across different build orientations and powder sources, evaluated for each grading strategy. The most pronounced distinctions were observed between specimens fabricated horizontally using waste L-PBF powder and those produced with virgin EB-PBF Arcam powder. Across all core geometries and grading methods, uniform, density-graded, and stress-graded, the specimens manufactured from virgin EB-PBF powder consistently exhibited reduced flexural yield strength and stiffness, yet demonstrated enhanced toughness in most cases.

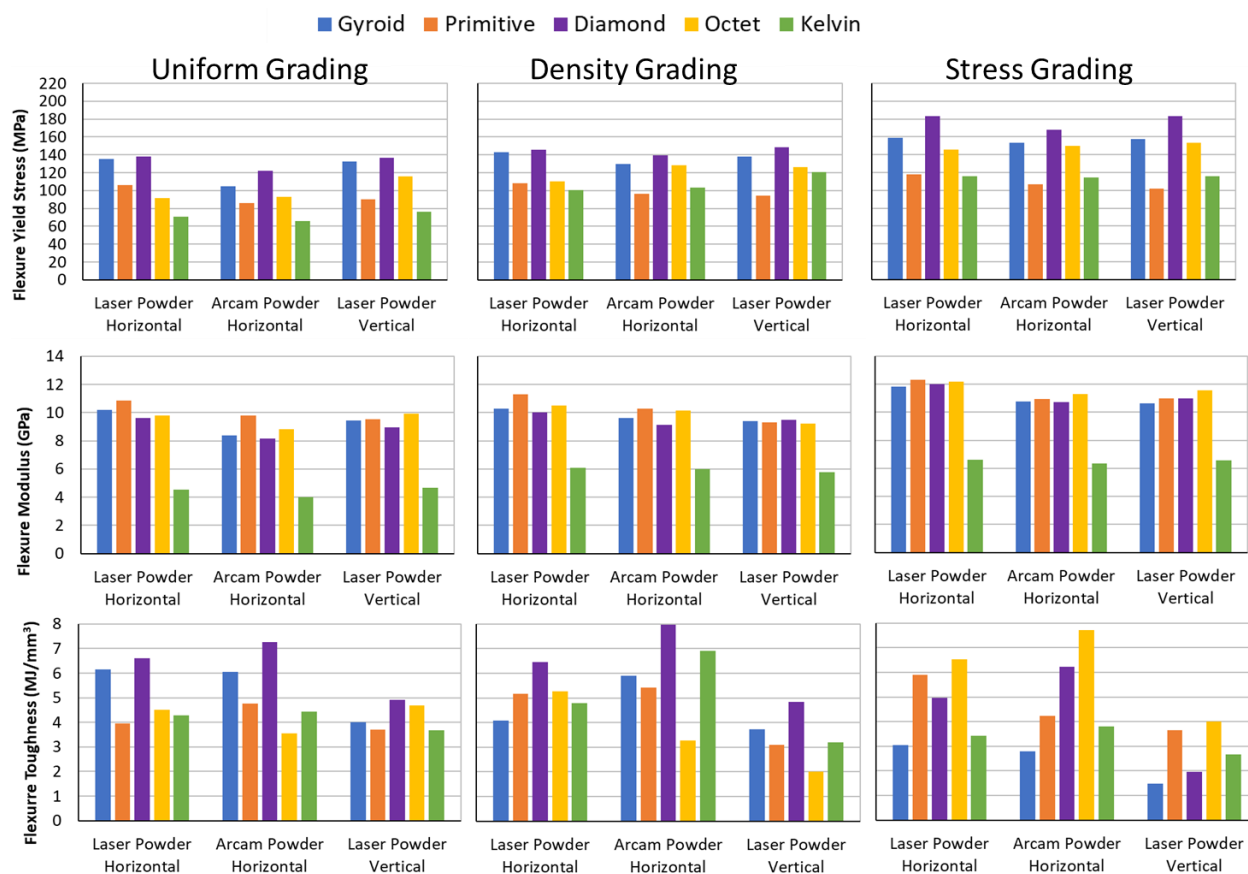


Figure 6.4. Comparison of flexural Yield Stress, Modulus, and Toughness properties for the specimens built horizontally and vertically with the waste L-PBF Ti-6Al-4V powder, and horizontally with the virgin EB-PBF Arcam Ti-6Al-4V powder.

These trends align with the tensile characterization results in Section 5.4.4, which revealed that the virgin EB-PBF Arcam powder produced material with lower ultimate strength but greater ductility. This mechanical profile shows that sandwich structures fabricated from virgin powder are more susceptible to early yielding under flexural loading but possess a higher tolerance for plastic deformation before failure. The increased toughness observed in these specimens can be attributed to their ability to absorb and dissipate strain energy more effectively, even as stiffness and strength metrics decline. The influence of build orientation further compounds these effects, as horizontal builds tend to promote more uniform layer bonding and reduced anisotropy, while vertical builds may introduce interlayer weaknesses that exacerbate early failure. The interplay between powder condition, build orientation, and grading strategy underscores the importance of integrated process-material design in optimizing the mechanical performance.

Specimens fabricated in the vertical orientation using waste L-PBF powder demonstrated reductions in both flexural yield strength and modulus relative to their horizontally built counterparts. However, these reductions were less pronounced than the disparities observed between different powder feedstocks. Notably, for strut-based lattice architectures such as the Octet and Kelvin configurations, certain vertically oriented specimens exhibited enhanced strength and stiffness compared to those built horizontally. This alludes that strut-dominated geometries are particularly sensitive to build orientation, and the directional dependence of grain morphology and interlayer bonding inherent to the EB-PBF process.

When comparing build orientation effects, particularly for the Diamond topology, vertically fabricated specimens exhibited a measurable reduction in flexural modulus relative to their horizontally built counterparts. Specifically, the modulus decreased by 7.2% for the uniform core, 5.3% for the density-graded core, and 8.8% for the stress-graded core. These reductions suggest

that anisotropy introduced during the layer-by-layer process, particularly in the vertical direction, can adversely affect mechanical performance with increased susceptibility to crack propagation under flexural loading.

In contrast, this trend does not extend to toughness, where vertically printed specimens consistently showed markedly lower values across all lattice types. The diminished toughness reflects a reduced capacity for energy absorption prior to failure, as evidenced by the accelerated failure progression observed in the stress-strain response. The vertical orientation appears to exacerbate strain localization and crack propagation, rendering the structure less tolerant to plastic deformation. Furthermore, the application of grading, whether density-based or stress-based, intensifies this vulnerability. Graded lattice specimens exhibited even more rapid damage evolution under flexural loading, likely due to stress concentration effects introduced by geometric transitions. These findings underscore the complex interplay between build orientation, lattice topology, and grading strategy in governing the mechanical behavior of EB-PBF cellular structures.

6.4 STRUCTURAL BEHAVIOR

The mechanical performance of additively manufactured lattice structures is intrinsically linked to the underlying microstructural characteristics imparted by the fabrication process. Variations in powder feedstock condition, build orientation, and thermal history during layer-by-layer deposition significantly influence grain morphology, phase distribution, and defect density, each of which governs the macroscopic structural behavior under flexural and tensile loading.

This section explores the interplay between microstructural features and the observed mechanical responses, including yield strength, stiffness, and toughness, across different lattice topologies and grading strategies. Particular attention is given to the anisotropic nature of the

microstructure resulting from directional solidification. These microstructural dependencies manifest in distinct failure modes, strain localization patterns, and energy absorption capacities, which are critical for the design and optimization of architected materials in load-bearing applications. By correlating microstructural observations with stress-strain behavior and fracture characteristics, this discussion aims to illuminate the mechanisms driving performance variability and to inform future strategies for tailoring mechanical properties through process control and material selection.

The high intrinsic quality of the EB-PBF processed material used in the sandwich structures was confirmed through mechanical characterization and microstructural evaluation, indicating minimal defect content such as porosity, lack of fusion, or inclusions. As a result, the observed structural behavior under flexural loading was predominantly governed by the geometric configuration of the core architectures rather than material imperfections. Failure mechanisms across the tested lattice topologies, including TPMS structures such as Gyroid, Primitive, and Diamond, as well as strut-based configurations like Octet and Kelvin, were strongly influenced by the stiffness and load distribution characteristics of the core geometries. TPMS-based sandwich beams consistently exhibited failure localized at the face sheets, with the Gyroid and Diamond structures primarily failing in tension at the lower face sheet, while the Primitive structure showed compressive failure at the upper face sheet. This directional dependence reflects the anisotropic stress states induced by the mixed-mode loading conditions and the inherent curvature and connectivity of the TPMS surfaces.

Conversely, strut-based lattices demonstrated core-dominated failure modes, characterized by localized buckling and crushing of struts near the loading interfaces. These failure patterns imply that strut geometries are more susceptible to instability under concentrated compressive loads,

particularly in regions of stress concentration adjacent to the test fixture. The implementation of density and stress grading strategies did not fundamentally alter the failure modes associated with each core geometry. However, these grading techniques contributed to a more optimized material distribution, enhancing the specific strength and delaying the onset of flexural yielding. By tailoring the internal architecture to better align with stress trajectories, graded lattices achieved improved mechanical efficiency without compromising structural integrity.

The flexural response of metallic sandwich structures incorporating TPMS lattice cores, specifically Gyroid, Diamond, and Primitive topologies, has garnered significant attention due to their superior mechanical performance compared to conventional honeycomb or strut-based architectures. Unlike traditional cellular designs, TPMS lattices inherently mitigate stress concentrations typically observed at nodal junctions and at the interface between the face sheet and the core. This is attributed to their continuous, smooth surface geometry and distributed porosity, which promote more uniform stress distribution under bending loads.

Among the TPMS variants examined, the Gyroid, Diamond, and Primitive configurations demonstrated enhanced resistance to crack initiation and propagation. Their interconnected porous topology disrupts linear crack trajectories, compelling crack paths to deviate and dissipate energy, thereby delaying catastrophic failure. This crack-arresting behavior was substantiated through post-mortem optical microscopy of failed specimens, wherein the Diamond structures exhibited consistent fracture patterns irrespective of the applied grading strategy, indicating that the intrinsic geometry played a dominant role over material distribution.

Ultimate failure in the graded Gyroid and Diamond sandwich panels was predominantly governed by tensile overload in the lower face sheet. Once the applied flexural stress exceeded the ultimate tensile strength of the fabricated Ti-6Al-4V material, rapid crack propagation ensued,

traversing vertically through the core and leading to structural collapse. Clearly, the implemented grading methodologies, whether density-based or stress-based, did not sufficiently account for the vulnerability of the lower face sheet under tensile loading. Consequently, they failed to arrest the dynamic stress wave transmission post-yield, underscoring a critical limitation in current grading paradigms for TPMS-based sandwich structures.

The most pronounced differences in mechanical performance between horizontally and vertically built Diamond-core specimens were attributed to the anisotropic microstructural orientation resulting from the build direction. As illustrated in Figure 6.5, the diagrams on the left and right delineate the build orientation and corresponding microstructural alignment with respect to the loading axis employed during the 4-point bend experiments. Metallographic cross-sections were extracted along both the width and thickness of the specimens, corresponding to the YZ and XY planes, to characterize the grain morphology, which revealed either columnar or equiaxed prior- β grain structures depending on the build orientation.

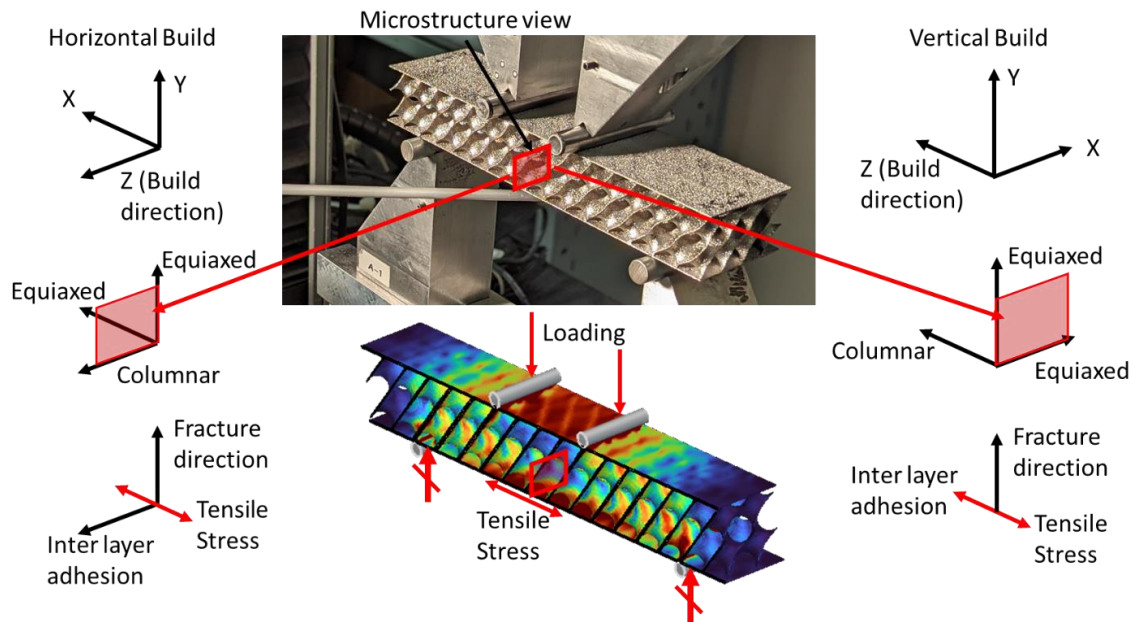


Figure 6.5. Diagram of the build direction and the resulting microstructural orientation relative to the 4-point bend loading condition.

Given that fracture propagation occurred perpendicular to the direction of tensile stress in the lower face sheet, the orientation of inter-layer boundaries relative to the crack path played a critical role in determining failure behavior. In horizontally built specimens, the tensile stress acted orthogonally to the columnar grain structure, compelling crack propagation to traverse across the grain boundaries. This orientation necessitated interaction with both the ductile β phase, enriched by Vanadium, and the stronger, more brittle α phase, stabilized by Aluminum. As the crack advanced, the columnar alignment imposed a tortuous path, either redirecting the crack along the β phase or requiring increased energy to penetrate the α phase, thereby enhancing resistance to fracture.

In contrast, vertically built specimens exhibited columnar grain alignment parallel to the tensile stress direction. This configuration facilitated a more direct crack trajectory, allowing the fracture to navigate around α phase columns and preferentially propagate through the β phase. The reduced path complexity and lower energy barrier in this orientation contributed to earlier onset of failure and diminished flexural strength. This anisotropic mechanical behavior is consistent with well-documented trends in additively manufactured tensile specimens, wherein components fabricated in the XY build plane typically exhibit superior mechanical properties compared to those printed along the Z-axis. Specifically, in EB-PBF processed Ti-6Al-4V alloys, the columnar prior- β grain microstructure tends to align with the build direction, which when coincident with the direction of applied tensile stress, results in diminished resistance to crack propagation through the material thickness.

This microstructural orientation significantly influences the flexural response of vertically built specimens. The alignment of elongated grains parallel to the tensile stress vector facilitates crack initiation and accelerates fracture progression, thereby manifesting as a reduced flexural

modulus, lower flexural yield strength, and premature failure onset relative to horizontally built counterparts. These observations underscore the critical role of grain morphology and orientation in governing the mechanical integrity of EB-PBF components under bending loads. Consequently, in the design and fabrication of load-bearing engineered structures, it is imperative to strategically consider the microstructural alignment of the printed material in relation to anticipated stress distributions. Optimal part orientation within the build chamber should be selected to ensure that the direction of maximum tensile stress intersects the grain structure in a manner that impedes crack propagation, preferably orthogonal to columnar grain growth to enhance structural resilience and prolong service life.

Chapter 7. CONCLUSION AND RECOMMENDATIONS

The conclusions presented in this chapter serves as the final synthesis of this investigation into the flexural behavior of Ti-6Al-4V sandwich structures, containing novel topology optimized lattice cores, produced by electron beam powder bed fusion (EB-PBF). Key findings, contributions, and implications of this investigation are encapsulated within, as well as a reflection on the research objectives. Additionally, this chapter provides the platform to discuss the limitations encountered throughout, and propose directions for future research. This study not only stands as a comprehensive piece of scholarship, but also paves the way for continued exploration and development in the field of additive manufacturing.

7.1 CONCLUSIONS

The goal of this work was to advance the state-of-the-art in metal additive manufacturing by exploring the flexural behavior of optimized Ti-6Al-4V sandwich structure produced using the EB-PBF process. This was accomplished by designing sandwich structures with constant thickness face sheets, and five different uniform thickness lattice core architectures to a lightweight relative density target of 10%. The lattice core geometries included three TPMS types, Gyroid, Primitive, and Diamond, and two traditional strut-types, Octet and Kelvin. Then two different optimization approaches were applied to functionally grade the core topologies with continuously variable thicknesses, while maintaining the 10% relative density target. The first grading scheme used a compliance minimization strategy based on the popular SIMP method, while the second used a stress minimization strategy based on a linear-static von Mises stress analysis.

Prior to fabrication of the sandwich specimens via EB-PBF from Ti-6Al-4V feedstock powder, manufacturing quality trials were required to guide the final configurations. However, the pedigree of the feedstock material was unknown, as it had come from the waste stream of a L-

PBF manufacturing site, and it had to be analyzed and tested prior to use. Through chemical, flowability, and morphological analysis, it was confirmed the powder was suitable for use in this research, but initial trials in the Arcam A2X machine resulted in smoke events. Through careful study of the physics involved in the smoking phenomenon, a novel approach was invented that involved adjusting the preheat strategy, and it resulted in a patent application. Upon successful completion of manufacturing quality trials, initial material tensile and compression tests demonstrated good strength and ductility.

Once the fifteen novel sandwich beam configurations were finalized, the EB-PBF build was designed such that the beams were oriented horizontally, with the face sheets oriented vertically for optimal quality. Four builds were completed, which delivered a minimum of three of each specimen; an error in the second lattice beam build resulted in only a partial completion of 10 specimens, minus the five stress graded beams. An additional build was completed with a set of specimens in the vertical orientation, and a final three smaller builds with virgin EB-PBF powder provided another set of horizontal specimens. The only post processing consisted of powder removal that required an average of one hour per specimen, and removal of support features. The 85 beam specimens were then tested in 4-point bend with third-span loading in accordance with ASTM C393 standard.

Analysis of the experimental data revealed that functional grading of lattice cores in an additively manufactured sandwich structure delivered higher flexural strength and damage resistance from conventional uniform density cores. The simpler von Mises stress minimization strategy out-performed the density-based compliance minimization strategy most commonly associated with topology optimization. Repeatability of the experimental data demonstrated the quality of the structures such that failure was driven by material or design, rather than by

manufacturing or process defects. Fractography and metallographic analyses concluded the material was of high quality with the lack of defects in the fracture surface and a fine $\alpha + \beta$ microstructure. The same trend in flexural stiffness was predicted by a linear-static FEA, though it overpredicted the performance with too much idealization. Also, the lack of a non-linear buckling analysis contributed to the overprediction.

Anisotropy was confirmed by the change in behavior between the horizontal and vertical build orientations. The alignment of the columnar microstructure with the primary tensile stress direction in the component proved to delay yielding and resist fracture. Horizontal specimens produced with the virgin EB-PBF Arcam powder showed slightly lower strength, but higher ductility and tolerated more flexural strain prior to ultimate failure. When that result was correlated with the initial and final tensile properties, it could be seen that the waste L-PBF powder, that produced many of the beam specimens, had started to become more oxygenated. That may have contributed to an increase in strength and a loss of ductility in specimens produced later in the investigation.

This research targeted aerospace industry applications by designing Ti-6Al-4V sandwich structures, to be weight optimized, based on the intended load case, using Design for Additive Manufacturing best practices. It was accomplished by designing a sandwich structure with sizing representative of a high temperature application such as a primary exhaust structure of a commercial turbofan engine. The face sheet thickness was selected based on commercially available AMS 4911, 0.032 inch sheet stock, with a lightweight core relative density of 10%, and a panel thickness of about 1 inch for high stiffness. Core designs were functionally graded based on a 4-point bend load case as defined by an industry standard ASTM C393 commonly used in aerospace.

Design of the beam specimens demonstrated many aspects of DfAM best practices. For example, a fillet, or blend radius was added to the face sheet-to-core interface as a means of reducing stress concentrations, and core web thicknesses were sized sufficiently to reduce the effect of surface roughness. Minimal feature size was based on the capability of the EB-PBF machine to realize them with adequate quality, and the core density was determined based on the ability to completely remove sintered powder from the sandwich internal cavity. For the overall build layout, build orientation was considered and optimized to have the face sheets always vertically oriented, which produced uniform surfaces without distortion. Support features were placed such that they could be removed, and sized for suitable heat conduction that mitigated heat accumulation. The successful fabrication of the large sandwich beam specimen builds, without failure caused by a build crash, is further evidence of the implementation of DfAM best practices.

In order to produce specimens using a common aerospace industry alloy such as Ti-6Al-4V, a feedstock source was identified from a L-PBF waste stream. The feedstock was first confirmed to be Ti-6Al-4V by chemical analysis, and the morphology was conditioned through powder handling processes to remove fines and oversized particles. A flowability analysis confirmed it was consistent with feedstock produced specifically for EB-PBF which meant it should behave similarly with the Arcam A2X powder distribution system. While smoke events with the powder during preheating posed a significant roadblock, this problem was mitigated through careful research and study of the physics behind the smoking phenomenon. This led to the invention of a novel preheating approach that used a step function current input that raised the temperature of the build plate at a slower rate. This allowed the powder to desensitize to the electro-static charge build up as the temperature increased. After which, tensile specimens were produced that confirmed the as printed material properties were comparable with wrought product Ti-6Al-4V.

Another accomplishment was the comparison of five lattice core topologies with three different grading methods applied, as well as observed anisotropy in build direction for each sandwich configuration. After an exhaustive study of available literature, there were two common lattice categories: 1) the traditional strut-type, and 2) a newer TPMS-type. Of the TPMS lattices, the three most common architectures were the Gyroid, Primitive, and Diamond, and they were compared against the Octet and Kelvin strut lattices. To differentiate from existing literature and provide meaningful contribution, these structures were integrated with face sheets into a sandwich structure, and functionally graded based on a 4-point bend simulation.

While a uniform grading served as a baseline control group, two different topology optimization methods were applied to the lattice core. The first approach employed a compliance minimization strategy using a commercial SIMP algorithm. The novelty of its implementation was to use a thickened lattice structure as the design space, and extracted the implicit density field that was mapped to the target geometry with upper and lower bounds tuned to provide a relative density of 10%. In contrast, a simpler yet more efficient method was to perform a linear-static FEA, then extract the implicit von Mises stress field and map it back to the target geometry, customizing the upper and lower bounds to also provide a relative density of 10%. Through these methods, the density- and stress-based approaches generated first-of-its-kind sandwich core geometries, graded in three dimensions, optimized for a combined mixed-mode compression and tension load condition.

Anisotropic effects were captured by producing four builds with all fifteen specimen configurations in the primary horizontal direction, and a fifth build was vertical, such that the specimen length was parallel to the build direction. From the 4-point bend experiments, a more brittle behavior was observed from those oriented vertically, such that failure occurred earlier in

some cases and damage was more rapid at lower flexural strains. This anisotropy was further observed when observing the microstructure. The orientation of the columnar prior- β grains in the primary loading direction meant that cracks could propagate more easily around them.

Sizing of the sandwich structures for 1) quality material, 2) topology driven flexural behavior, and 3) complete powder removal was also accomplished. To determine adequate sandwich lattice core sizing, manufacturing quality trials were conducted with meso-scale lattices at various densities. Quality material through the thickness was confirmed by the microstructure analysis that showed a fine $\alpha + \beta$ expected of Ti-6Al-4V. Also, consistency of the 4-point bend load-deflection data validated that the observed behaviors were not defect driven. Powder removal was initially verified by visual inspection, and subsequently by the mass/density analysis. Further evidence that complete powder removal was obtained, was that no residual loose powder exited the specimens during flexural experiments. The consistency of load-deflection results would have been adversely affected had compacted sintered powder been present and influencing the internal load distributions.

Experimental repeatability was accomplished after determining the flexural behavior empirically, by industry standard ASTM C393 4-point bend, with a minimum of three of each specimen configuration. The first step in achieving this was during specimen design, which was completed according to guidance provided by ASTM C393. Although, since AM lattice structures are still very new, some modifications were needed for the lattice core topologies in order to meet the intent of the standard. Then the specimens were tested in 4-point bend with third-span loading to provide adequate bending deformation. For repeatability, 3-4 copies of each of the 15 specimen designs were built horizontally, one set of 15 was printed vertically, and another set was printed

with virgin EB-PBF powder. This made for a total of 85 specimens tested, with no outliers or unexplained behaviors.

This investigation established several key learnings and insights into the design and performance of additively manufactured sandwich structures using EB-PBF technology. First, it was demonstrated that preheat process parameters can be strategically adjusted to accommodate waste L-PBF powder, enabling sustainable reuse without compromising build quality. In terms of structural optimization, stress minimization consistently outperformed compliance minimization, leading to more mechanically efficient designs. Functional grading of the lattice cores was shown to delay the onset of flexural yielding, allowing the structures to achieve higher loads and sustain greater strains before failure.

Among the various core geometries evaluated, TPMS-based designs exhibited superior resistance to core buckling, with the Diamond topology achieving the highest overall performance regardless of grading strategy. Finally, the research underscored a critical design-for-additive-manufacturing (DfAM) principle: when considering the build orientation of a component, the direction of the expected peak tensile stresses should avoid alignment with the build direction. Doing so can increase mechanical performance by preventing unfavorable microstructural orientation.

Several impactful contributions were made by this study to the field of metal additive manufacturing, particularly in the context of lightweight sandwich structure design for aerospace. It demonstrated the utility of repurposing waste powder from the L-PBF process, showing that recycled feedstock can still yield good mechanical properties in EB-PBF. The research also documented how variations in powder feedstock influence mechanical behavior, providing valuable insights for sustainable material use. In terms of applicability, the work presented the

first known comparison between functionally graded metal AM sandwich structures and established ASTM standards, bridging the gap between experimental design and industry benchmarks.

From a design simplicity standpoint, the analysis revealed that traditional density-based topology optimization methods are not well-suited for lattice structures, whereas stress grading offers a more straightforward and effective alternative. It also showcased the implementation of a linear-static FEA method to predict performance trends in graded sandwich structures, streamlining the design process. Finally, this work emphasized quality by delivering material-driven behavior understanding with a strong focus on design for additive manufacturing (DfAM), reinforcing the importance of aligning material characteristics with structural performance.

7.2 RECOMMENDATIONS

The findings presented throughout this research offer a foundation for advancing both the scientific understanding and practical implementation of additively manufactured sandwich structures. Building on the experimental results, computational analyses, and microstructural evaluations, this section outlines targeted recommendations for future research, design optimization, and industrial application. These recommendations aim to address observed limitations, refine current methodologies, and guide the development of more robust, efficient, and scalable solutions. By integrating insights from material behavior, process parameters, and structural performance, the following proposals seek to inform next-generation strategies in metal additive manufacturing and lattice-based structural design.

The economics of powder bed fusion processes present significant challenges when relying exclusively on virgin powder or incurring costs associated with disposing of unused or degraded feedstock. Given the high expense of virgin metal powders and the environmental and financial

implications of waste management, it is increasingly impractical for manufacturers to maintain a single-use approach to powder materials. This reality underscores the urgent need for further research into feedstock material behavior, particularly in areas such as powder reuse strategies, degradation mechanisms during repeated cycles, and the development of robust quality control methods for recycled powders. Advancing knowledge in these domains will not only improve the cost-efficiency and sustainability of PBF technologies but also support broader adoption across industries where resource optimization is critical. Thus, more research for feedstock material is recommended in the following areas.

- Case studies on the effects of heavily cycled powder on actual components.
- Identification of waste streams from other processes for repurposing powder.
- Alternative metal powder waste and process parameter optimization.
- Processes or parameters that restore non-idealized powder into quality material.
- In EB-PBF, non-line-of-sight powder removal methods to facilitate removal of sintered powder from lattice structures, particularly TPMS types.
- Performance effects of post-process smoothing methods on lattice sandwich structures in static and fatigue conditions.

Design for additive manufacturing (DfAM) plays a pivotal role, not only in ensuring the successful integration of AM technologies within industrial settings, but also in shaping the quality and relevance of academic research that drives the field forward. Effective DfAM principles enable engineers and researchers to fully leverage the unique capabilities of AM processes, such as geometric freedom, material efficiency, and functional integration. Moreover, thoughtful design directly influences the reliability, repeatability, and scalability of experimental outcomes, which are essential for generating insights that can be translated into real-world applications.

Recognizing this dual importance, the following recommendations focus on optimizing the design and build of experiments to enhance both practical implementation and scholarly contribution.

- Consider the rotation of the unit cells to eliminate either detrimental or beneficial alignment with the test apparatus and loading direction (e.g. TPMS Primitive or Octet topologies).
- Expand on the effects of build orientation anisotropy with respect to loading at a component level, not just the coupon level.
- Investigate specimens of different overall relative densities with the same optimization methods to build an analytical model (e.g. Gibson-Ashbey) for the behavior of these topologies.
- Behavior of larger, more open lattice unit cells, perhaps with thicker walls, is needed to make the structures more applicable to a broader range of industries.
- Ergonomics of EB-PBF build handling requires consideration during build layout, as tall builds of sintered metal powder are heavy, which leads to a higher risk of damage during extraction and the potential for negatively affecting experimental results.

This research demonstrated the effectiveness of implicit modeling combined with linear-static analysis in simulating and optimizing lattice core sandwich structures, showcasing their potential for accurate and efficient structural evaluation. These methods enabled the exploration of complex geometries and mechanical behaviors with reduced computational overhead, contributing valuable insights into the design and performance of advanced lightweight materials. However, to further advance the science of numerical modeling in this domain, additional steps are necessary. These include incorporating nonlinear and dynamic analysis to capture real-world loading conditions, refining mesh generation techniques for intricate lattice geometries, and integrating multi-physics

simulations that account for thermal, dynamic, and manufacturing-induced effects. Expanding the modeling framework in the following directions will enhance predictive accuracy, broaden applicability, and support the development of more resilient and functional lattice-based structures.

- Consider meshless methods for optimization.
- Consider face sheet grading in combination with lattice core grading.
- Explore phase-field (boundary modification) optimization for grading of sandwich structure geometry.
- Consider non-linear core buckling in lattice core and face sheet thickness grading.
- Consider methods of homogenization for larger models and graded core densities.
- Consider elastic-plastic material model in the optimization scheme.
- Incorporate the statistical probability of the effect of various defect populations and anisotropy in lattice functional grading.
- Expand the optimization and functional grading to multiple standard test methods.
- Apply thermo-mechanical optimization techniques for heat flux and strength in lattice sandwich core.
- Apply optimization methods for cyclical conditions to enhance fatigue performance.

The excellent repeatability observed in the 4-point bend experimental data underscores the importance of meticulous planning, precise execution, and rigorous attention to detail in achieving reliable and meaningful results. This consistency validates the experimental setup and highlights the robustness of the procedures used. Despite the success in repeatability, there remain opportunities to deepen the understanding of lattice sandwich structure behavior. Enhancements such as incorporating high-resolution strain mapping, conducting tests under varied loading conditions, and integrating real-time deformation tracking could have provided richer data and

more nuanced insights. These additions would not only strengthen the interpretation of mechanical performance, but also contribute to refining predictive models and informing future design strategies.

- Employment of an audible sensor to identify and capture when small fractures occur during loading which are not visually observable.
- Include datum or locating features outside the test region of interest in order to obtain a more robust method of alignment between the specimens and the experimental apparatus.
- Inclusion of non-contact strain monitoring such as digital image correlation (DIC) to measure face sheet and core displacement.
- Measure the effect of strain rate on lattice sandwich structures optimized for static conditions.
- Study the effects of heat conduction and flux through functionally graded structures.
- Expand flexural testing of optimized lattice sandwich structures to fatigue experiments.

The recommendations presented here are intended to guide future exploration and practical advancement in the field. By highlighting areas for methodological refinement, encouraging interdisciplinary collaboration, and advising pathways for technological innovation, this section aims to extend the relevance and applicability of the research. These proposals are not only grounded in the findings of the current study, but also reflect broader opportunities for impact across academic and industrial domains. As researchers and practitioners build upon this work, it is hoped that these recommendations will serve as a catalyst for continued progress and discovery.

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APPENDIX A

As-received Ti-6Al-4V Chemical Analysis (2/7/2020)

- A sample of Ti-6Al-4V powder was analyzed using a Rigaku NEX CG Energy Dispersive X-Ray Fluorescence Spectrometer.
- Instrument fundamental parameter quantitative analysis was performed.
- Standardization was performed using SRM654a, Titanium Alloy, 6Al-4V a certified standard in the form of a disk, 31 mm in diameter and 6.4 mm thick used for application in x-ray spectrometric analysis.
- The sample was analyzed three times with the powder mixed between runs to redistribute the components.

Element	Run #1 (mass %)	Run #2 (mass %)	Run #3 (mass %)	Avg (mass %)	Standard SRM 654A		
					Element	Run	Certified Value
Titanium (Ti)	88.6	90.5	90.2	89.8	Titanium (Ti)	89.4	89.4
Aluminum (Al)	5.02	5.07	5.11	5.07	Aluminum (Al)	6.4	6.3
Vanadium (V)	6	4.09	4.35	4.81	Vanadium (V)	3.4	3.9
Iron (Fe)	0.144	0.143	0.137	0.141	Iron (Fe)	0.179	<0.2
Zirconium (Zr)	0.0946	0.105	0.116	0.105	Zirconium (Zr)	0.083	0
Chromium (Cr)	0	0.0902	0.0818	0.0573	Chromium (Cr)	0.223	0
Silicon (Si)	0.025	0.0258	0.0257	0.0255	Silicon (Si)	0.193	0
Tin (Sn)	0.0202	0.0231	0.0244	0.0226			

Titanium Powder ONH Results (2/7/20)				
TI-64 D	Oxygen%	Nitrogen%	Hydrogen%	Carbon%
Run 1	0.138	0.315	0.00778	0.0125
Run 2	0.141	0.318	0.00832	0.0131
Run 3	0.139	0.319	0.0081	0.0119
Avg	0.139	0.317	0.00807	0.0125

Flowability Analysis of Donated 'Oversized' Ti-6AL-4V powder (As received)

							Mass	50 grams	(scale used to measure 50g samples was off)	
Time (seconds) for 50g of powder to pass through a hall flow meter							Correctior	0.798403	(real mass was 67.8g)	
Test Run	Sample 1	Sample 2	Sample 3	Flow rate, corrected (s)						
1	34.01	34.62	31.91	27.15	27.64	25.48				
2	34.41	34.68	32.15	27.47	27.69	25.67				
3	34.17	35.03	32.05	27.28	27.97	25.59				
4	33.71	35.11	32.18	26.91	28.03	25.69				
5	33.71	34.98	31.71	26.91	27.93	25.32				
				Avg	27.15	27.85	25.55	Method 1:		
								FR _H =	27 s/50g	
Measurement of the time required for 50 g of a metallic powder to flow through the orifice of a calibrated funnel of standardized dimensions.										
ASTM B213 Standard Test Method for Flow Rate of Metal Powders										
ISO 4490 Metallic powders - Determination of flow rate by means of a calibrated funnel (Hall Flowmeter)										
ASTM B212 Standard Test Method for Apparent Density of Free Flowing Metal Powders Using the Hall Flowmeter Funnel										
ISO No. 3923-1 Metallic powders - Determination of apparent density - Part 1 Funnel method										
Calculate Flow Rate and Apparent Density										
Measured mass was incorrect, 50g = 67.8g										
times above require a correction factor of 0.798										
Correction = (expected time)/(actual time) of chinese emery										
exected	40		40							
actual	50.1		38.2							
corr.	0.798403		1.04712							
	(mass incorrect)		(correct mass)							

Test 2/6/2021: After PRS and sieved 4 times, 3 times since first build attempt

							Mass	50.0 grams	(used new scale from previous measurement)
Time (seconds) for 50g of powder to pass through a hall flow meter							Correction	1.058901	
Test Run	Calibration	Sample 1	Sample 2	Sample 3		Flow rate, corrected (s)			
1	37.55	21.92	21.78	22.16		23.21	23.06	23.47	
2	38.22	22.12	22.13	22.36		23.42	23.43	23.68	
3	37.95	22.08	22.13	22.45		23.38	23.43	23.77	
4	37.72	22.13	22.15	22.36		23.43	23.45	23.68	
5	37.88	22.26	22.28	22.88		23.57	23.59	24.23	
	37.775				Avg	23.40	23.40	23.76	Method 1: FR _H = 24 s/50g
Measurement of the time required for 50 g of a metallic powder to flow through the orifice of a calibrated funnel of standardized dimensions.									
ASTM B213 Standard Test Method for Flow Rate of Metal Powders									
ISO 4490 Metallic powders - Determination of flow rate by means of a calibrated funnel (Hall Flowmeter)									
ASTM B212 Standard Test Method for Apparent Density of Free Flowing Metal Powders Using the Hall Flowmeter Funnel									
ISO No. 3923-1 Metallic powders - Determination of apparent density - Part 1 Funnel method									
Calculate Flow Rate and Apparent Density									
Correction = (expected time)/(actual time) of chinese emery									
exected	40								
actual	37.775								
corr.	1.05890139								

Feedstock Powder Handling Weight Records

Batch 1	Weight	Delta			Notes
Extracted	112	lbs	50.80	kg	(included 3-small container weights not in follow-on measurements)
PRS 1	103	lbs	9	46.72	kg (weight recorded after 1st build attempt)
PRS 2	100	lbs	3	45.36	kg
PRS 3	97	lbs	3	44.00	kg
PRS 4	94	lbs	3	42.64	kg
PRS 5	92	lbs	2	41.73	kg Weight following successful Line build, B1
PRS 6	91.5	lbs	0.5	41.50	kg Weight following unsuccessful lattice v1 build, B2 (catastrophic @ 5.3 mm)
PRS 7	90	lbs	1.5	40.82	kg Weight following unsuccessful lattice v1 build, B3 (filament failure @ 12.9 mm)
Batch 2	Weight	Delta			Notes
Extracted	90	lbs	40.82	kg	
PRS 3	74.5	lbs	15.5	33.79	kg
Add Batch 1	159.5	lbs	-85	72.35	kg Weight following partially successful lattice v1 build, B4 (lattice successful, tensile coupons failed)
PRS 4	155.5	lbs	4	70.53	kg Weight following successful lattice v2 build, B5
PRS 5	138	lbs	17.5	62.60	kg Weight following successful lattice v3 (beam) build, B6
Batch 3	Weight	Delta			Notes
Extracted	112	lbs	50.80	kg	
PRS 3	98	lbs	14	44.45	kg
Add Batch 2	214	lbs	-116	97.07	kg Weight following partial lattice v3 build #2, B7, and added powder
PRS 4	N/A		N/A		kg No weight recorded for lattice v4 build #3 failure B8
PRS 5	178	lbs	36	80.74	kg Weight following partial lattice v3 build #4, B9
PRS 6	161	lbs	17	73.03	kg Weight following successful lattice v3 build #5, B10
PRS 7	141.5	lbs	19.5	64.18	kg Weight following successful lattice vertical, B11
PRS 8	134	lbs	7.5	60.78	kg Weight following successful lattice v3, B12 (coupon build)

B0.1 (1/30/21)



B0.2 (1/30/21)



B0.3 (2/6/21)



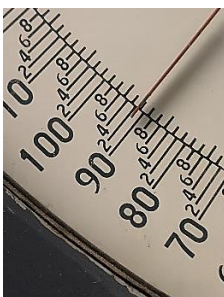
B0.4 (2/6/21)



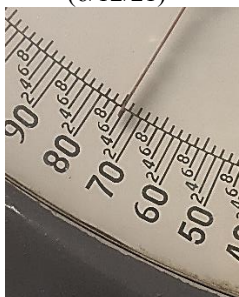
B1 (2/28/21)



B2 (5/2/21)



B3 (6/6/21)

Add post PRS
(6/12/21)

B4 (6/13/21)



B5 (6/28/21)



B6 (10/17/21)

Add post PRS
(1/8/22)

B9 (2/6/22)



B10 (2/20/22)



B11 (3/27/22)

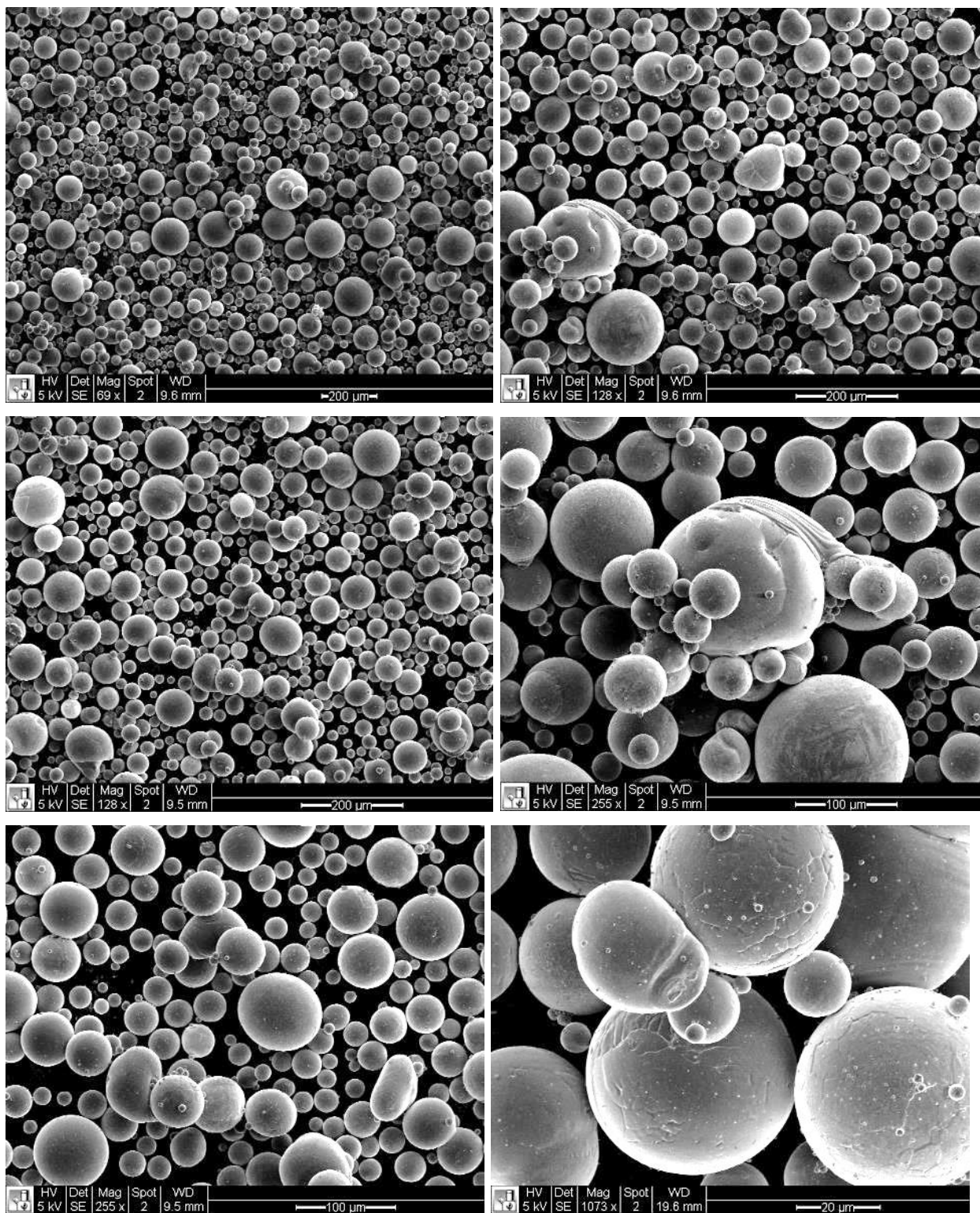


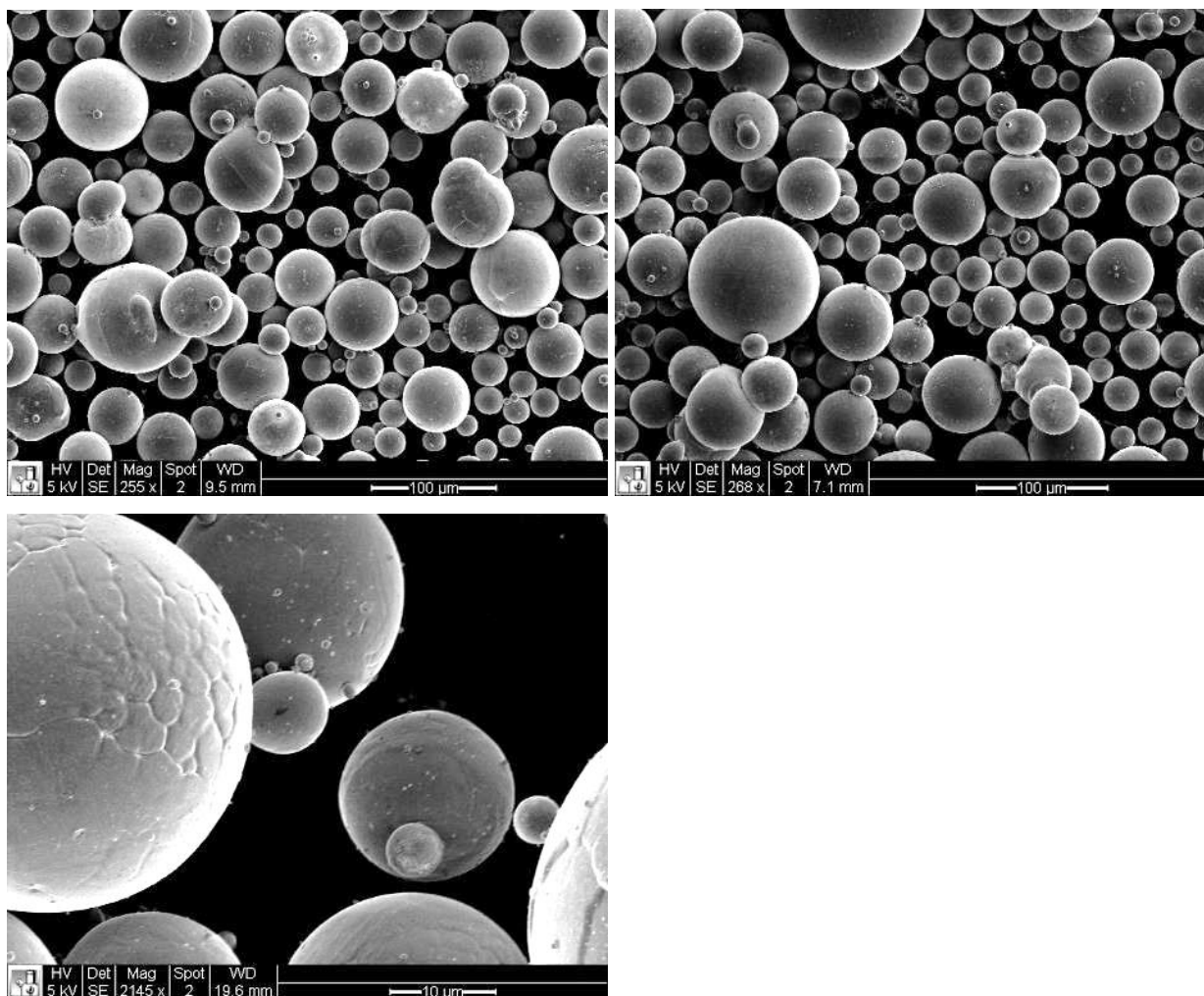
Example Method for Weighing:



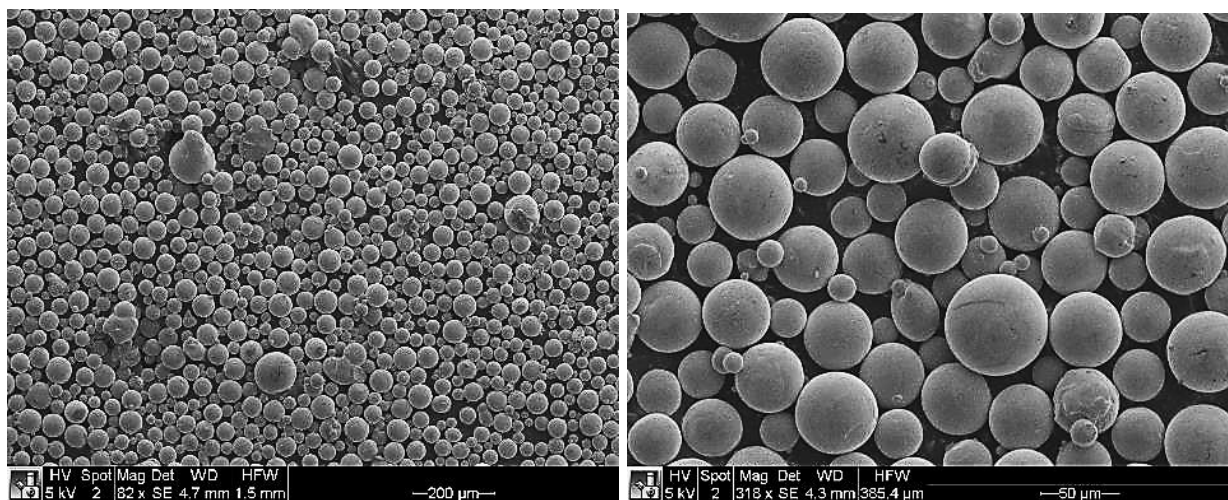
Morphology Analysis

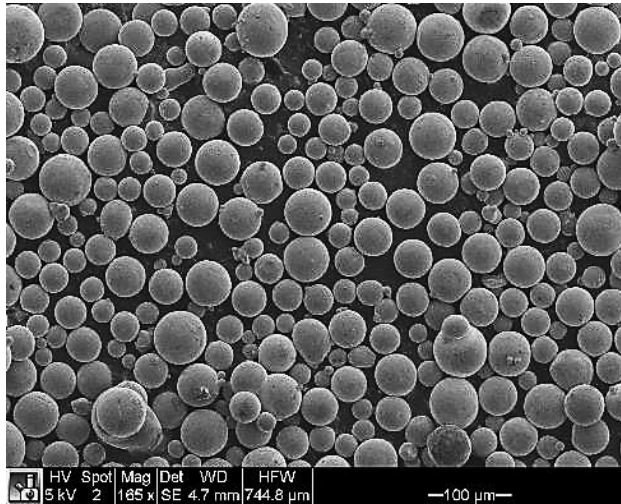
SEM Images of powder feedstock (as-received condition)



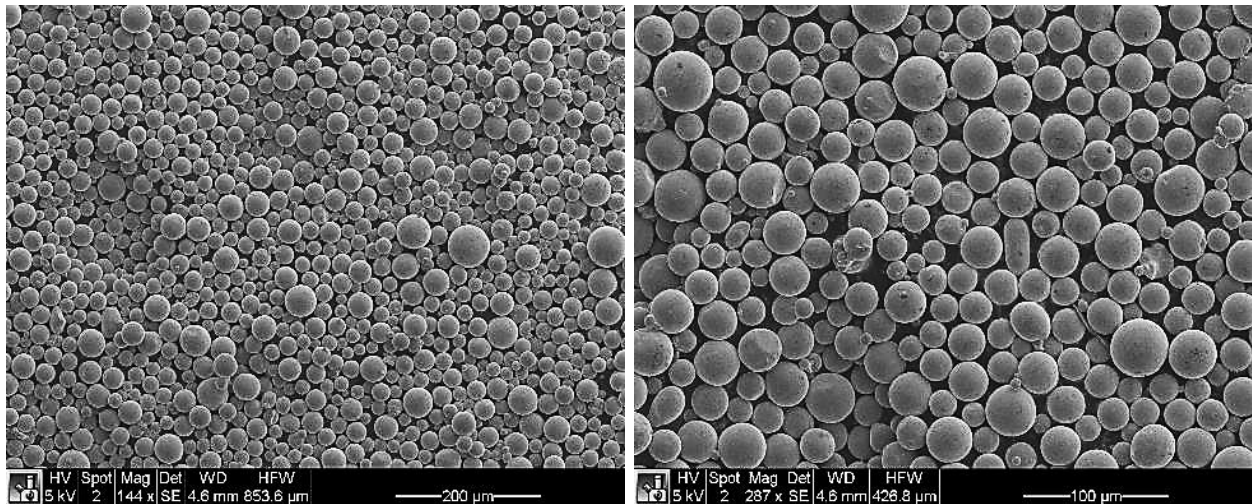


SEM Images of powder feedstock (PRS and sieved 4x condition)





SEM Images of powder feedstock (fines removed by PRS < 45 μm)



MATLAB Analysis code

(File formatting is Sample_LowerBounds_UpperBounds_Sensitivity(if exists, otherwise just blank).png)

```
% Temp particle distribution
clear all, close all, clc
% Location, sample types, etc.
loc = 'C:\Users\fig\FindCirc\';
type = ['AR', 'SV', 'FN']; % AS-RECEIVED; SIEVED; FINES
% images = {'SV1_165x.png', 'SV2_318x.png', 'SV3_82x.png'};
% images = {'AR1_69x.png', 'AR2_128x.png', 'AR3_128x.png'};
images = {'FN1_144x.png', 'FN2_287x.png'};
nfls = length(images);
fls = cell(1, nfls);
for i = 1:nfls
```

```

    % Construct the full path for each image file
    fullFilePath = [loc images{i}]; % This creates a full file path as a
string

    % Use dir to get the file information and store it in the cell array
    fls{i} = dir(fullFilePath);
end
ebf = 65; % Edge buffer
% ardi = [];
% svdi = [];
fndi = [];
for i = 1:nfls
    fnm = fls{i}.name;
    pimg = imread([loc fnm]);
    figure(i)
    imshow(pimg)
    % As recieved
    % lowerBounds = [6, 10, 6]
    % upperBounds = [80, 60, 75]
    % Sensitivity = [0.9, 0.9, 0.9]
    % Sieved
    % lowerBounds = [10, 10, 6];
    % upperBounds = [100, 150, 50];
    % Sensitivity = [0.91, 0.915, 0.9];
    % Fines
    lowerBounds = [6, 6]
    upperBounds = [40, 70]
    Sensitivity = [0.9, 0.91]
    [centers, radii, metric] = imfindcircles(pimg(1:(end-ebf),:),...
        [lowerBounds(i)
upperBounds(i)], 'Sensitivity', Sensitivity(i));
    hold on
    viscircles(centers(1:end,1:2), radii(1:end), 'EdgeColor', 'r');
    set(gcf, 'WindowState', 'maximized'); % Maximize the current figure
    drawnow;
    pimg = frame2im(getframe(gcf));

    fileName = [num2str(lowerBounds(i)) '_' num2str(upperBounds(i)) '_'
num2str(Sensitivity(i)) '_' images{i}];
    imwrite(pimg, strcat(loc, fileName));

    % Get user input for scale
    [xtmp, ytmp] = ginput;
    xlg = length(xtmp);
    % Continue until 2X selection
    while xlg ~= 2
        [xtmp, ytmp] = ginput;
        xlg = length(xtmp);
    end
    sblg = input('What is the scale reading (um)? ');
    npx = max(xtmp)-min(xtmp); sf = sblg/npx; % Scale factor
    dtmp = 2*radii(1:end)*sf;
    if fnm(1:2) == 'AR'
        ardi = vertcat(ardi, dtmp);
    elseif fnm(1:2) == 'SV'
        svdi = vertcat(svdi, dtmp);

```

```

elseif fnm(1:2) == 'FN'
    fndi = vertcat(fndi,dtmp);
else
    % Error handling (not implemented)
end
end
save('psd_extract_test.mat','fndi');
load('psd_extract_test.mat');
%% Plot statistics
close all
n_bins = 60;
fsz = 20;
% Remove outliers (default = 3 STDEV)
% ardir = rmoutliers(ardi);
% svdir = rmoutliers(svdi);
% fndir = rmoutliers(fndi);
circleCount = 0;
D10 = 0;
D50 = 0;
D90 = 0;
plotTitle = '';
if exist('ardi', 'var')
    figure(1)
    % subplot(1,3,1)
    histfit(ardi,n_bins)
    %histfit(ardir,n_bins)
    title('As-Received Powder','FontSize',fsz)
    xlabel('Diameter, \mum','FontSize',fsz)
    ylabel('Count','FontSize',fsz)
    xlim([0,150]); ylim([0,1000]);
    set(gca,'FontSize',fsz)
    D10 = prctile(ardi, 10);
    D50 = prctile(ardi, 50);
    D90 = prctile(ardi, 90);
    hold on;
    plot([D10, D10], ylim, 'r--', 'LineWidth', 2);
    plot([D50, D50], ylim, 'r--', 'LineWidth', 2);
    plot([D90, D90], ylim, 'r--', 'LineWidth', 2);
    hold off;
    circleCount = length(ardi);
    plotTitle = 'AR';
end
if exist('svdi', 'var')
    % figure(2)
    histfit(svdi,n_bins)
    % histfit(svdir,n_bins)
    title('Sieved Powder','FontSize',fsz)
    xlabel('Diameter, \mum','FontSize',fsz)
    ylabel('Count','FontSize',fsz)
    set(gca,'FontSize',fsz)
    xlim([0,105]); ylim([0,450]);
    D10 = prctile(svdi, 10);
    D50 = prctile(svdi, 50);
    D90 = prctile(svdi, 90);
    hold on;
    plot([D10, D10], ylim, 'r--', 'LineWidth', 2);

```

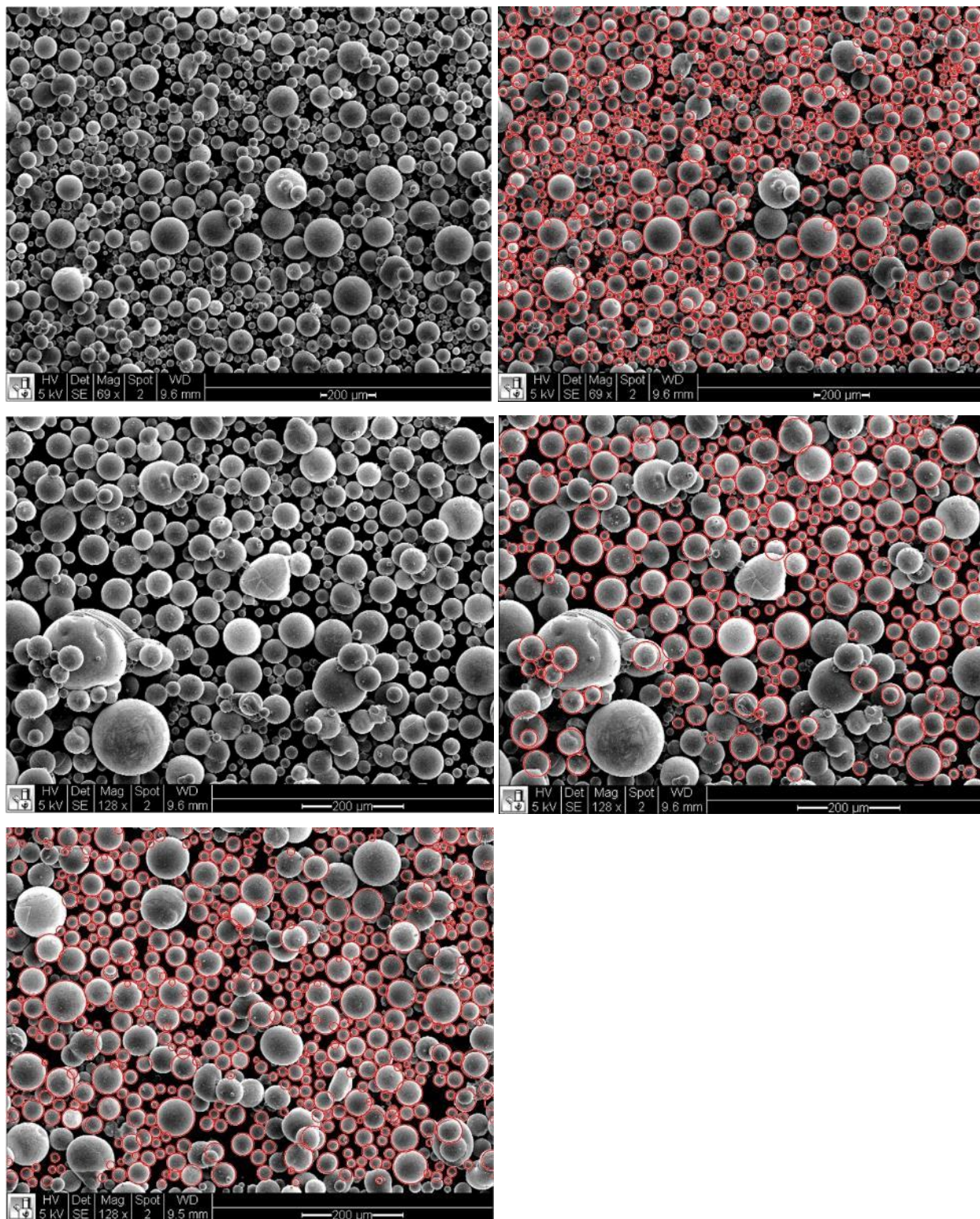
```

    plot([D50, D50], ylim, 'r--', 'LineWidth', 2);
    plot([D90, D90], ylim, 'r--', 'LineWidth', 2);
    hold off;
    circleCount = length(svdi);
    plotTitle = 'SV';
end
if exist('fndi', 'var')
    % subplot(1,3,3)
    histfit(fndi,n_bins)
    title('Powder Fines','FontSize',fsz)
    xlabel('Diameter, \mum','FontSize',fsz)
    ylabel('Count','FontSize',fsz)
    set(gca,'FontSize',fsz)
    xlim([0,50]); ylim([0,450]);
    D10 = prctile(fndi, 10);
    D50 = prctile(fndi, 50);
    D90 = prctile(fndi, 90);
    hold on;
    plot([D10, D10], ylim, 'r--', 'LineWidth', 2);
    plot([D50, D50], ylim, 'r--', 'LineWidth', 2);
    plot([D90, D90], ylim, 'r--', 'LineWidth', 2);
    hold off;
    circleCount = length(fndi);
    plotTitle = 'FN';
end
set(gcf, 'WindowState', 'maximized'); % Maximize the current figure
drawnow;

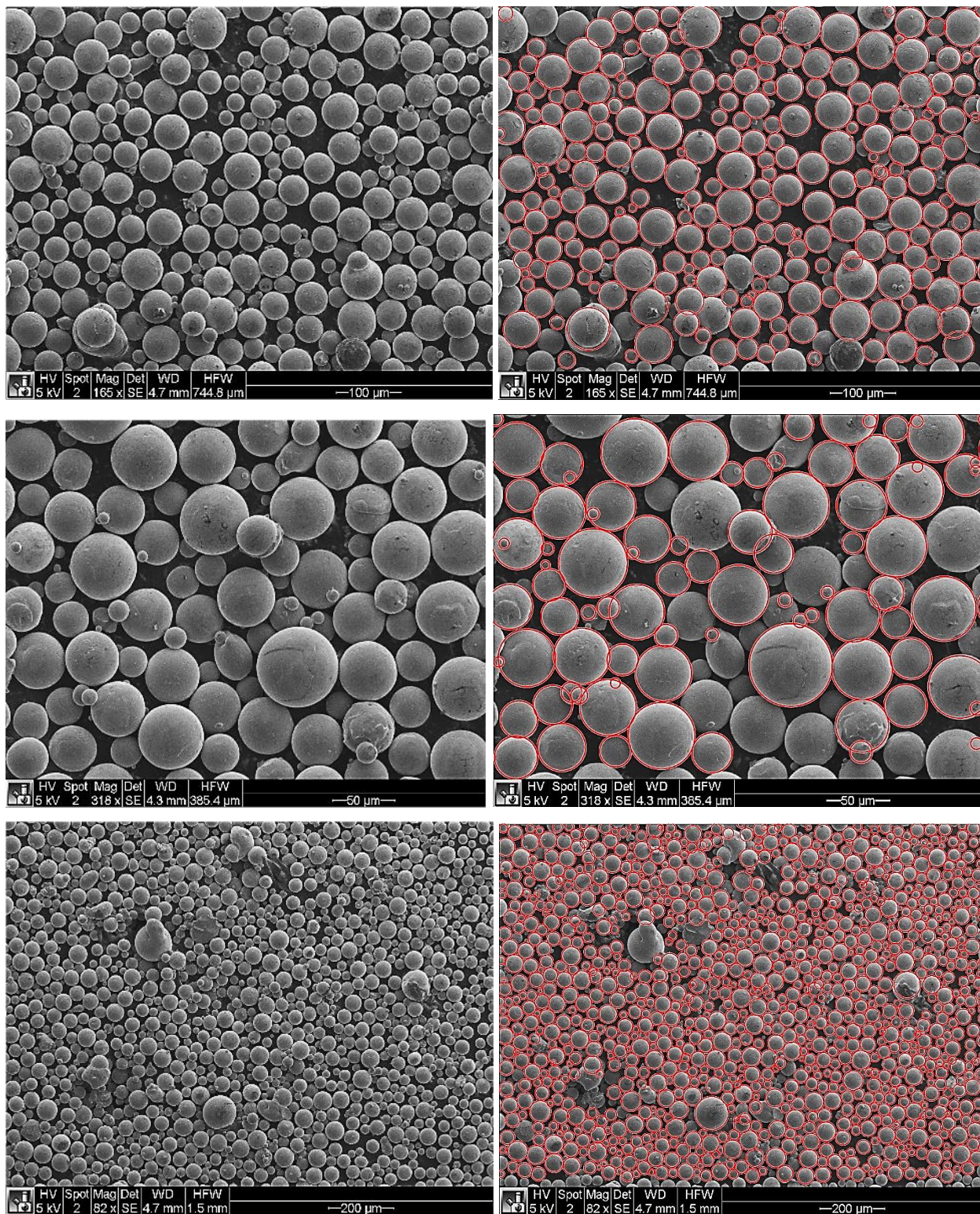
%Significant figures after decimal point
sigFigs = 2;
pimg = frame2im(getframe(gcf));
sigFigsFormat = sprintf('%.%df', sigFigs);
fileName = [num2str(sprintf(sigFigsFormat, D10)) '_'
num2str(sprintf(sigFigsFormat, D50)) '_' num2str(sprintf(sigFigsFormat, D90))
 '_' num2str(circleCount) '_' plotTitle '.png'];
imwrite(pimg, strcat(loc, fileName));
%% Test
%{
pimg = imread('test_particles.png');
figure(1)
imshow(pimg)
[centers, radii, metric] = imfindcircles(pimg,[5.0 180],'Sensitivity',0.90);
hold on
viscircles(centers(4:end,1:2),radii(4:end),'EdgeColor','r');
% Scale factor
sblg = 200; npx = 143; sf = sblg/npx;
%sblg = 200; npx = 190; sf = sblg/npx;
dtmp = 2*radii(4:end)*sf;
% Plot PSD
figure(2)
histfit(dtmp,20)
%}

```

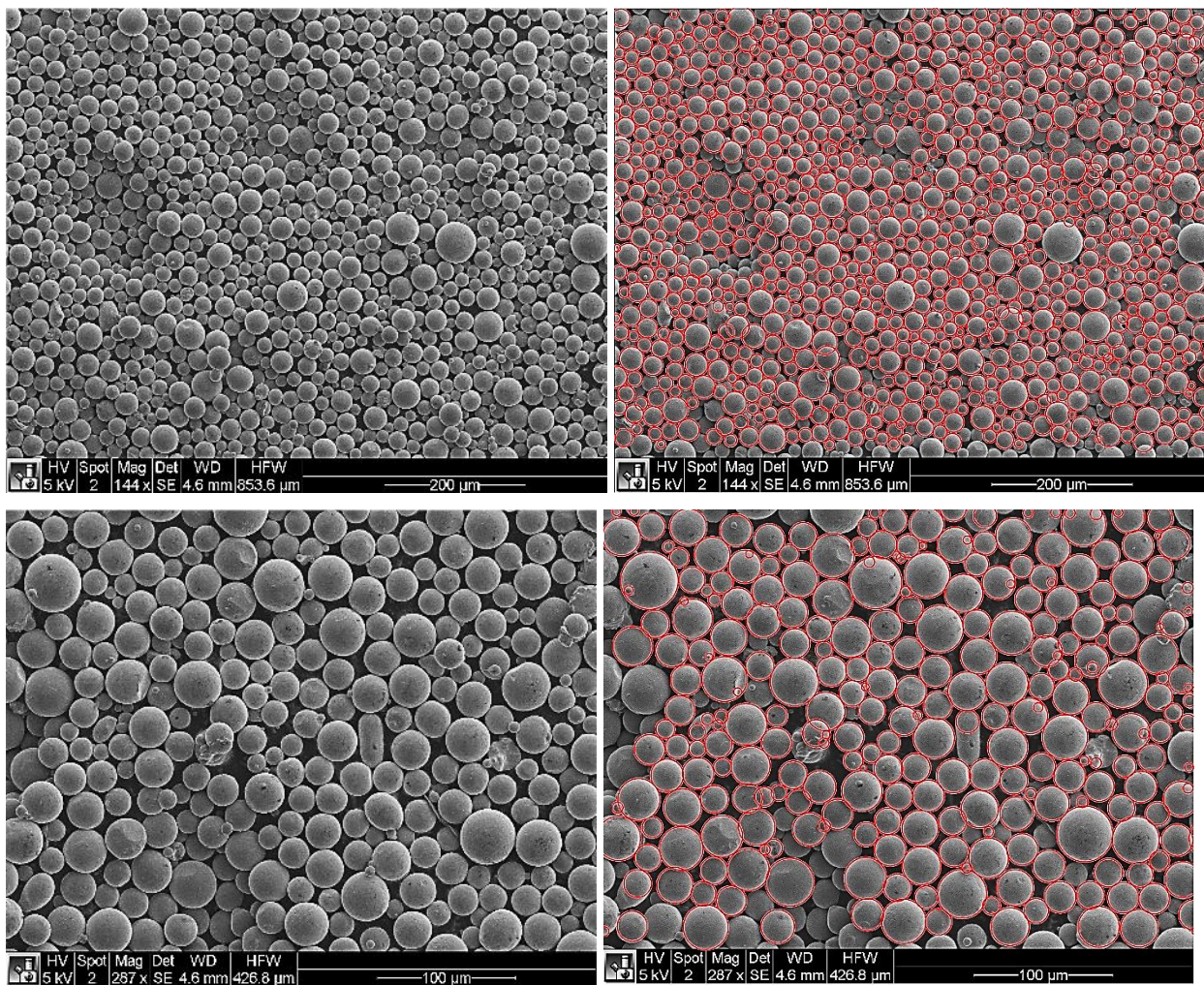

Example image analysis files (as-received)



Example image analysis files (after PRS and sieved)



Example image analysis files (fines removed by PRS < 45 μm)



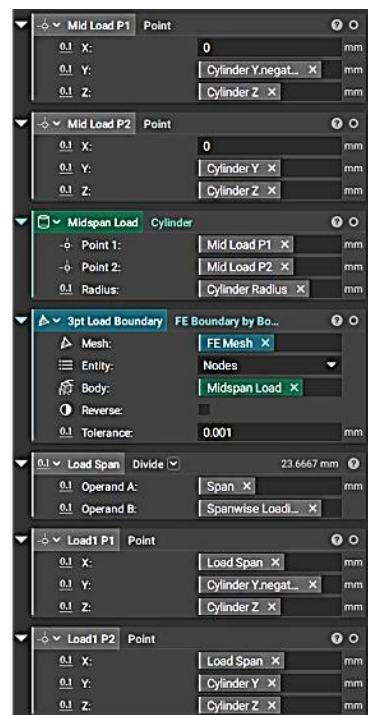
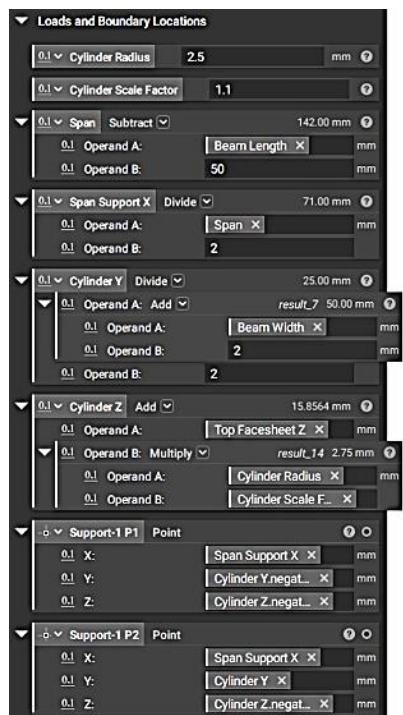
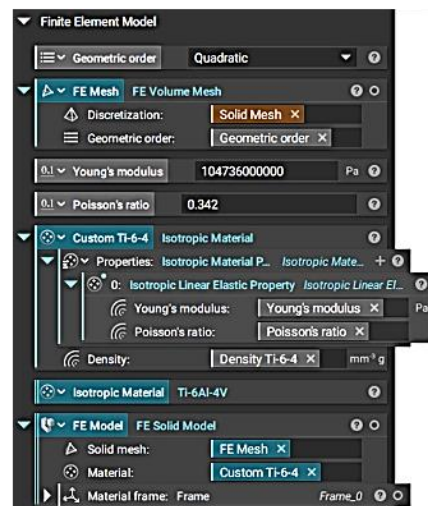
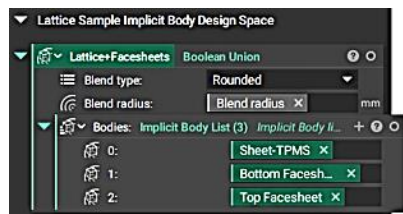
APPENDIX B

Expanded Blocks of nTopology Model Workflow:

Uniform Graded Model

The image displays a collection of nTopology model workflow blocks for a Uniform Graded Model, organized into six panels:

- Beam Geometric Variables:**
 - Beam Length: 192 mm
 - Beam Width: 48 mm
 - Beam Height: 25.4 mm
 - Lattice Thickness: 1.1 mm
 - Facesheet thickness: 0.8128 mm
 - Infill Lattice Width: Subtract, 46.60 mm (Operand A: Beam Width, Operand B: 1.4)
 - Infill Lattice Length: Subtract, 190.60 mm (Operand A: Beam Length, Operand B: 1.4)
 - Unit Cell Size: 24, 24, 25.4 mm
 - Density TI-6-4: 0.00443 mm³ g
 - Blend radius: .8 mm
- Model Components:**
 - Design Space Volume:** Box, Center point: 0, 0, 0 mm; Length: Infill Lattice Len... mm; Width: Infill Lattice Width mm; Height: Beam Height mm.
 - Top Facesheet Z:** Divide, 13.1064 mm (Operand A: Add, result_9 26.2128 mm; Operand B: 2)
 - Bottom Facesheet Z:** Divide, -13.1064 mm (Operand A: Add, result_9_copy 26.2128 mm; Operand B: -2)
 - Top Facesheet Center:** Point, X: 0 mm, Y: 0 mm, Z: Top Facesheet Z mm
 - Bottom Facesheet C...:** Point, X: 0 mm, Y: 0 mm, Z: Bottom Facesheet Z mm
- Top Facesheet Details:**
 - Top Facesheet:** Box, Center point: Top Facesheet ... mm; Length: Beam Length mm; Width: Beam Width mm; Height: Facesheet thick... mm.
 - Bottom Facesheet:** Box, Center point: Bottom Facesheet... mm; Length: Beam Length mm; Width: Beam Width mm; Height: Facesheet thick... mm.
 - DS+Facesheet Solid:** Boolean Union, Blend type: Rounded, Blend radius: Blend radius mm; Bodies: Implicit Body List (3) Implicit Body Li... (0: Design Space V... mm, 1: Bottom Facesheet... mm, 2: Top Facesheet mm).
- Measures:**
 - Facesheet:** Mass Properties from Body, Body: Top Facesheet mm, Density: Density TI-6-4 mm³ g, Relative error: 0.01.
 - Lattice Volume:** Mass Properties from Body, Body: Design Space V... mm, Density: Density TI-6-4 mm³ g, Relative error: 0.01.
 - Sample Full Solid:** Mass Properties from B..., Body: DS+Facesheet... mm, Density: Density TI-6-4 mm³ g, Relative error: 0.01.
 - Sample:** Mass Properties from Body, Body: Lattice+Facesheet... mm, Density: Density TI-6-4 mm³ g, Relative error: 0.01.
 - Sample Rel. Density:** Divide, 0.1623 (Operand A: Divide, result_L1 0.0007 mm³ g; Operand B: Sample Full Sol... mm³ g)
- Lattice Type Variables:**
 - Strut Lattice:** Infill Volume Lattice, Body: Design Space V... mm, Lattice type: Octet, Cell size: Unit Cell Size mm, Rotation: 0, 0, 0, Beam thickness: Lattice Thickness mm.
 - Strut Lattice:** Mass Properties from Body, Body: Strut Lattice mm, Density: Density TI-6-4 mm³ g, Relative error: 0.01.
 - Strut Rel. Density:** Divide, 0.0231 (Operand A: Divide, result_3_copy_c... 0.0001 mm³ g; Operand B: Lattice Volume... mm³ g)
 - Skeletal-TPMS:** TPMS, Body: Design Space V... mm, Cell size: Unit Cell Size mm, Mid-surface offset: 0 mm, Fill type: Lidlind.
 - Skeletal-TPMS:** Mass Properties from Body, Body: Skeletal-TPMS mm, Density: Density TI-6-4 mm³ g, Relative error: 0.01.
 - Skel-TPMS Rel. Den...:** Divide, 0.3861 (Operand A: Divide, result_3_copy 0.0017 mm³ g; Operand B: Lattice Volume... mm³ g)
- Sheet-TPMS Details:**
 - Sheet-TPMS:** Walled TPMS, Body: Design Space V... mm, Cell size: Unit Cell Size mm, Approx. thickness: Lattice Thickness mm, Mid-surface offset: 0 mm, Fill type: Diamond.
 - Sheet-TPMS:** Mass Properties from Body, Body: Sheet-TPMS mm, Density: Density TI-6-4 mm³ g, Relative error: 0.01.
 - Sheet-TPMS Rel. De...:** Divide, 0.1049 (Operand A: Divide, result_3 0.0005 mm³ g; Operand B: Lattice Volume... mm³ g)



Load Point 1 Cylinder

Point 1: Load1 P1 X mm

Point 2: Load1 P2 X mm

Radius: Cylinder Radius X mm

4pt P1 Boundary FE Boundary by Body

Mesh: FE Mesh X

Entity: Nodes

Body: Load Point 1 X

Reverse: ☐

Tolerance: 0.001 mm

Load2 P1 Point

X: Load Span.nega... X mm

Y: Cylinder Y.negat... X mm

Z: Cylinder Z X mm

Load2 P2 Point

X: Load Span.nega... X mm

Y: Cylinder Y X mm

Z: Cylinder Z X mm

Load Point 2 Cylinder

Point 1: Load2 P1 X mm

Point 2: Load2 P2 X mm

Radius: Cylinder Radius X mm

4pt P2 Boundary FE Boundary by Body

Mesh: FE Mesh X

Entity: Nodes

Body: Load Point 2 X

Reverse: ☐

Tolerance: 0.001 mm

3pt Load Vector Vector (0.00, 0.00, -30. N)

X: 0 N

Y: 0 N

Z: Total Load X N

4pt Load Vector Vector (0.00, 0.00, -15. N)

X: 0 N

Y: 0 N

Z: Divide result_12 -15,000.00 N

Operand A: Total Load X N

Operand B: 2

Point Map 4pt Load Vis

Points: Point List (2)

0: Load Point 1 bo... X mm

1: Load Point 2 bo... X mm

Values: Vector List (2)

0: 4pt Load Vector X N

1: 4pt Load Vector X N

3Pt Applied Load Force

Boundary: 3pt Load Bound... X

Vector: 3pt Load Vector X N

Frame: Optional

4Pt Applied Load 1 Force

Boundary: 4pt P1 Boundary X

Vector: 4pt Load Vector X N

Frame: Optional

4Pt Applied Load 2 Force

Boundary: 4pt P2 Boundary X

Vector: 4pt Load Vector X N

Frame: Optional

Support 1 Restraint Displacement Restr...

Boundary: Support1 Bound... X

Ux: 0 mm

Uy: 0 mm

Uz: 0 mm

Rx: 0 deg

Ry: Optional deg

Rz: 0 deg

Frame: Optional

Support 2 Restraint Displacement Restr...

Boundary: Support2 Bound... X

Ux: Optional mm

Uy: 0 mm

Uz: 0 mm

Rx: 0 deg

Ry: Optional deg

Rz: 0 deg

Frame: Optional

Static Analysis Check

Third Point Loading = 6, Quarter Point Loading = 4

Spanwise Loading: 6

Total Load: -30000 N

Static Analysis Static Analysis

Model: FE Model X

Load case: Boundary Conditio... Boundary Con...

0: 4Pt Applied Loa... X

1: 4Pt Applied Loa... X

2: Support 1 Restr... X

3: Support 2 Restr... X

Post Process Analysis

get stress points Von Mises Stress Po...

Analysis result: Static Analysis X

Step number: Optional

Ti-6-4 UTS: 9.935e+08 Pa

Fail criteria > UTS Filter (11237)

List: get stress point... X mm

Values: get stress point... X Pa

Comparison: Greater Than or Equal t

Reference: Ti-6-4 UTS X Pa

Ratio points > UTS Divide 0.0158

Operand A: Fail criteria > UT... X

Operand B: get stress point... X

Ti-6-4 YTS: 9.134e+08 Pa

Yield criteria > YTS Filter (14483)

List: get stress point... X mm

Values: get stress point... X Pa

Comparison: Greater Than or Equal t

Reference: Ti-6-4 YTS X Pa

Ratio points > YTS Divide 0.0204

Operand A: Yield criteria > Y... X

Operand B: get stress point... X

Integer_0: Empty

Notes Scalar List (5)

0: 0

1: 0

2: 0

3: 0

4: 0

Compliance Minimization Density Graded Model

The image displays a CAD software interface with several panels and components for a Compliance Minimization Density Graded Model.

Beam Geometric Variables

- Beam Length: 192 mm
- Beam Width: 48 mm
- Beam Height: 25.4 mm
- Lattice Thickness: 4 mm
- Facesheet thickness: 0.8128 mm
- Infill Lattice Width: Subtract, 46.60 mm
 - Operand A: Beam Width
 - Operand B: 1.4
- Infill Lattice Length: Subtract, 190.60 mm
 - Operand A: Beam Length
 - Operand B: 1.4
- Unit Cell Size: 24, 24, 25.4 mm
- Density Ti-6-4: 0.00443 mm³ g
- Blend radius: .8 mm

Model Components

- Design Space Volume** (Box)
 - Center point: 0, 0, 0 mm
 - Length: Infill Lattice Len...
 - Width: Infill Lattice Width
 - Height: Beam Height
- Top Facesheet Z** (Divide)
 - Operand A: Add, result_9, 26.2128 mm
 - Operand B: Beam Height
- Bottom Facesheet Z** (Divide)
 - Operand A: Add, result_9_copy, 26.2128 mm
 - Operand B: Facesheet thick...
- Top Facesheet Center** (Point)
 - X: 0 mm
 - Y: 0 mm
 - Z: Top Facesheet Z
- Bottom Facesheet C...** (Point)
 - X: 0 mm
 - Y: 0 mm
 - Z: Bottom Faces...
- Top Facesheet** (Box)
 - Center point: Top Facesheet ...
 - Length: Beam Length
 - Width: Beam Width
 - Height: Facesheet thick...
- Bottom Facesheet** (Box)
 - Center point: Bottom Facesh...
 - Length: Beam Length
 - Width: Beam Width
 - Height: Facesheet thick...
- DS+Facesheet Solid** (Boolean Union)
 - Blend type: Rounded
 - Blend radius: Blend radius
 - Bodies: Implicit Body List (3)
 - 0: Design Space V...
 - 1: Bottom Facesh...
 - 2: Top Facesheet

Measures

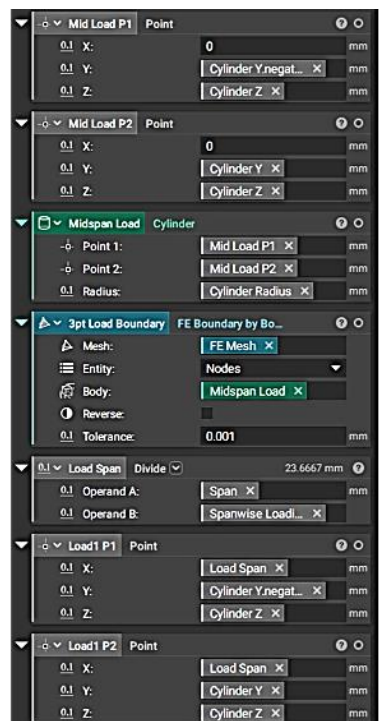
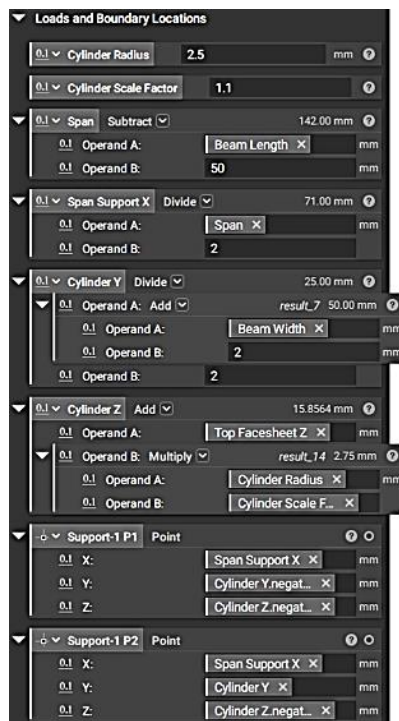
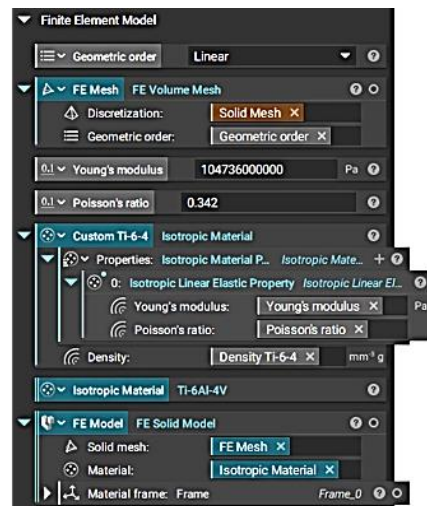
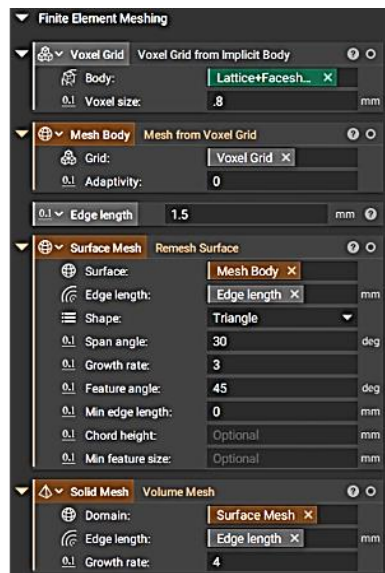
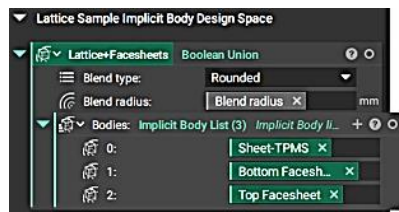
- Facesheet** (Mass Properties from Body)
 - Body: Top Facesheet
 - Density: Density Ti-6-4
 - Relative error: 0.01
- Lattice Volume** (Mass Properties from Body)
 - Body: Design Space V...
 - Density: Density Ti-6-4
 - Relative error: 0.01
- Sample Full Solid** (Mass Properties from B...)
 - Body: DS+Facesheet ...
 - Density: Density Ti-6-4
 - Relative error: 0.01
- Sample** (Mass Properties from Body)
 - Body: Lattice+Facesh...
 - Density: Density Ti-6-4
 - Relative error: 0.01
- Sample Rel. Density** (Divide)
 - Operand A: Divide, result_1, 0.0019 mm³ g
 - Operand B: Sample Full Sol...

Lattice Type Variables

- Strut Lattice** (Infill Volume Lattice)
 - Body: Design Space V...
 - Lattice type: Octet
 - Cell size: Unit Cell Size
 - Rotation: 0, 0, 0
 - Beam thickness: Lattice Thickness
- Strut Lattice** (Mass Properties from Body)
 - Body: Strut Lattice
 - Density: Density Ti-6-4
 - Relative error: 0.01
- Strut Rel. Density** (Divide)
 - Operand A: Divide, result_3_copy_c..., 0.0012 mm³ g
 - Operand B: Strut Lattice ma...
- Skeletal-TPMS** (TPMS)
 - Body: Design Space V...
 - Cell size: Unit Cell Size
 - Mid-surface offset: 0
 - Fill type: Lidoid
- Skeletal-TPMS** (Mass Properties from Body)
 - Body: Skeletal-TPMS
 - Density: Density Ti-6-4
 - Relative error: 0.01
- Skel-TPMS Rel. Den...** (Divide)
 - Operand A: Divide, result_3_copy, 0.0017 mm³ g
 - Operand B: Skeletal-TPMS...

Sheet-TPMS

- Sheet-TPMS** (Walled TPMS)
 - Body: Design Space V...
 - Cell size: Unit Cell Size
 - Approx. thickness: Lattice Thickness
 - Mid-surface offset: 0
 - Fill type: Diamond
- Sheet-TPMS** (Mass Properties from Body)
 - Body: Sheet-TPMS
 - Density: Density Ti-6-4
 - Relative error: 0.01
- Sheet-TPMS Rel. De...** (Divide)
 - Operand A: Divide, result_L3, 0.0017 mm³ g
 - Operand B: Sheet-TPMS.m...



Load Point 1 Cylinder

Point 1: Load1 P1 X mm

Point 2: Load1 P2 X mm

Radius: Cylinder Radius X mm

4pt P1 Boundary FE Boundary by Body

Mesh: FE Mesh X

Entity: Nodes

Body: Load Point 1 X

Reverse: ☐

Tolerance: 0.001 mm

Load2 P1 Point

X: Load Span.nega... X mm

Y: Cylinder Y.nega... X mm

Z: Cylinder Z X mm

Load2 P2 Point

X: Load Span.nega... X mm

Y: Cylinder Y X mm

Z: Cylinder Z X mm

Load Point 2 Cylinder

Point 1: Load2 P1 X mm

Point 2: Load2 P2 X mm

Radius: Cylinder Radius X mm

4pt P2 Boundary FE Boundary by Body

Mesh: FE Mesh X

Entity: Nodes

Body: Load Point 2 X

Reverse: ☐

Tolerance: 0.001 mm

3pt Load Vector Vector (0.00, 0.00, -30. N)

X: 0 N

Y: 0 N

Z: Total Load X N

4pt Load Vector Vector (0.00, 0.00, -15. N)

X: 0 N

Y: 0 N

Z: Divide result_12 -15,000.00 N

Operand A: Total Load X N

Operand B: 2

Point Map 4pt Load Vis

Points: Point List (2) Point List_4

0: Load Point 1.bo... X mm

1: Load Point 2.bo... X mm

Values: Vector List (2) Vector List_0

0: 4pt Load Vector X N

1: 4pt Load Vector X N

3Pt Applied Load Force

Boundary: 3pt Load Bound... X

Vector: 3pt Load Vector X N

Frame: Optional

4Pt Applied Load 1 Force

Boundary: 4pt P1 Boundary X

Vector: 4pt Load Vector X N

Frame: Optional

4Pt Applied Load 2 Force

Boundary: 4pt P2 Boundary X

Vector: 4pt Load Vector X N

Frame: Optional

Support 1 Restraint Displacement Restr...

Boundary: Support1 Bound... X

Ux: 0 mm

Uy: 0 mm

Uz: 0 mm

Rx: 0 deg

Ry: Optional deg

Rz: 0 deg

Frame: Optional

Support 2 Restraint Displacement Restr...

Boundary: Support2 Bound... X

Ux: 0 mm

Uy: 0 mm

Uz: 0 mm

Rx: 0 deg

Ry: Optional deg

Rz: 0 deg

Frame: Optional

Static Analysis Check

Third Point Loading = 6, Quarter Point Loading = 4

Spanwise Loading: 6

Total Load: -30000 N

Static Analysis Static Analysis

Model: FE Model X

Load case: Boundary Conditio... Boundary Con... +

0: 4Pt Applied Loa... X

1: 4Pt Applied Loa... X

2: Support 1 Restr... X

3: Support 2 Restr... X

Post Process Analysis

get stress points Von Mises Stress Pol...

Analysis result: Static Analysis X

Step number: Optional

TI-6-4 UTS: 9.935e+08 Pa

Fail criteria > UTS Filter (0)

List: get stress point... X mm

Values: get stress point... X Pa

Comparison: Greater Than or Equal t

Reference: TI-6-4 UTS X Pa

Ratio points > UTS Divide

Operand A: Fail criteria > UT... X

Operand B: get stress point... X

TI-6-4 YTS: 9.134e+08 Pa

Yield criteria > YTS Filter (0)

List: get stress point... X mm

Values: get stress point... X Pa

Comparison: Greater Than or Equal t

Reference: TI-6-4 YTS X Pa

Ratio points > YTS Divide

Operand A: Yield criteria > Y... X

Operand B: get stress point... X

Integer_0: Empty

Notes Scalar List (5)

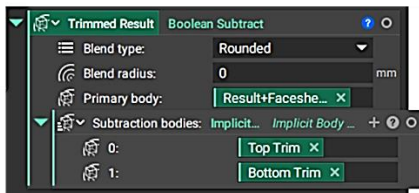
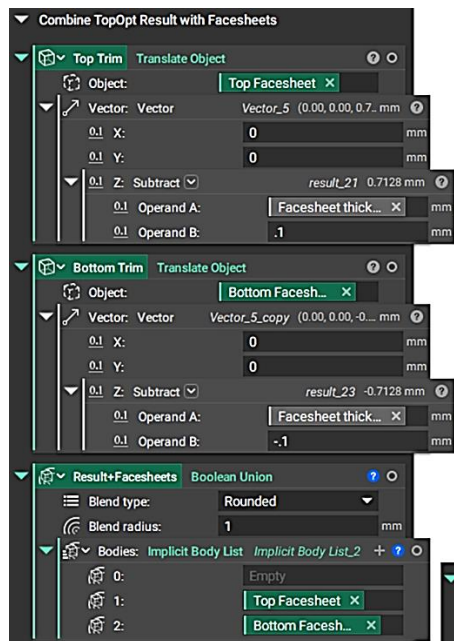
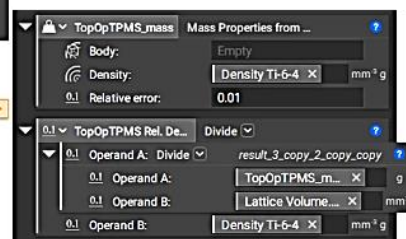
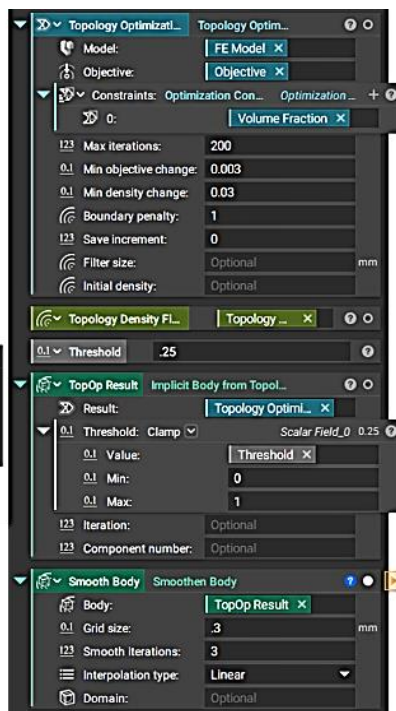
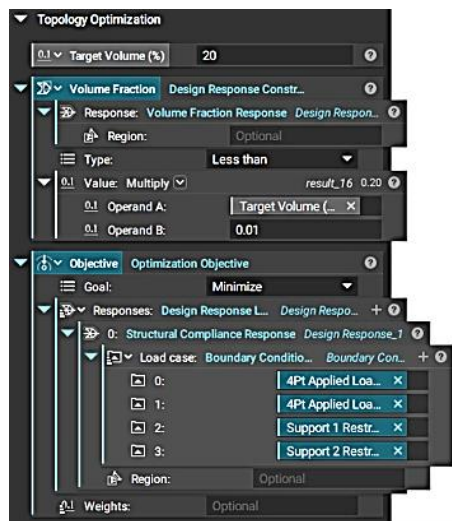
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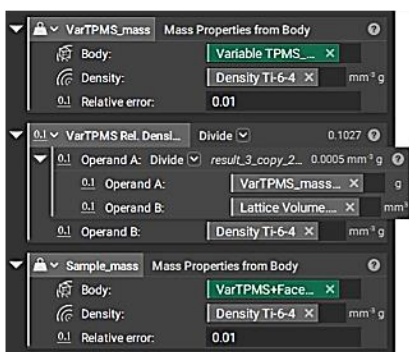
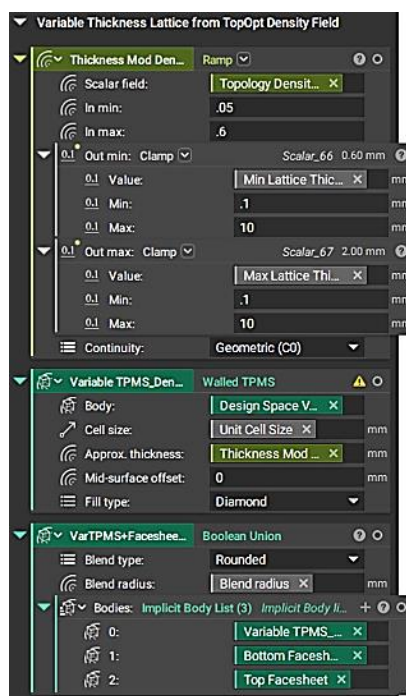
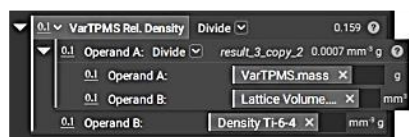
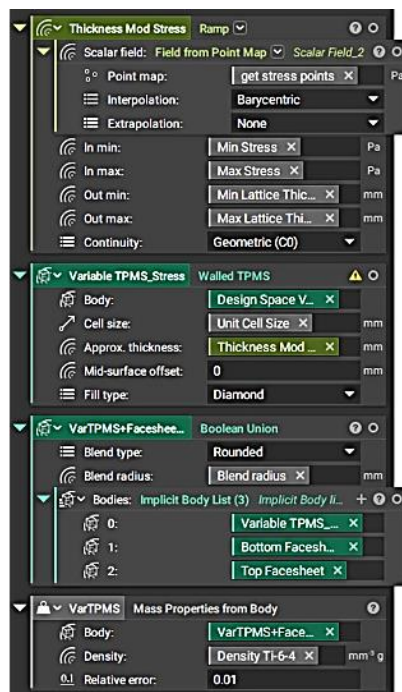
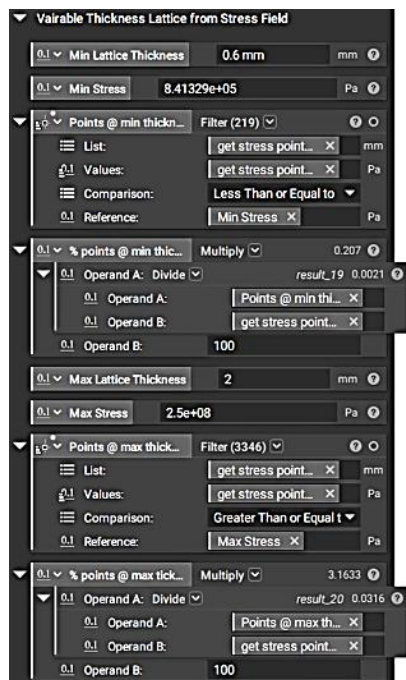
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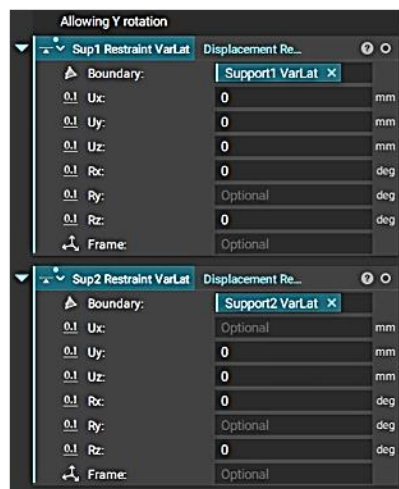
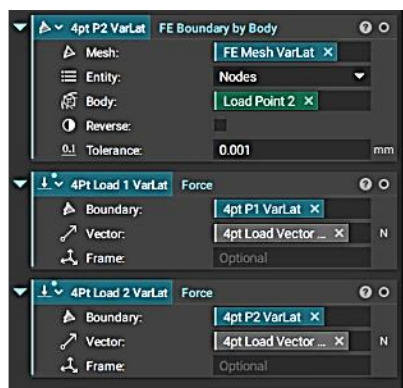
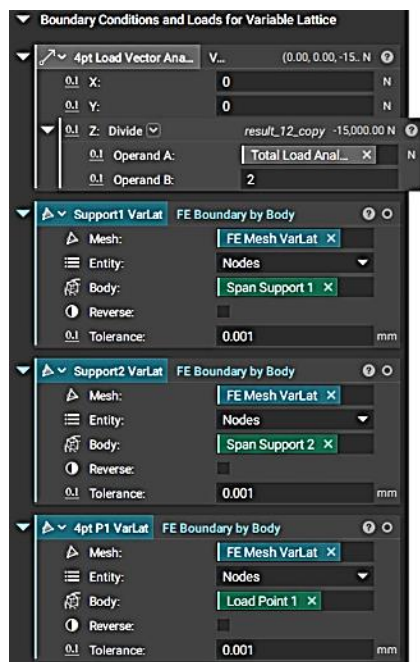
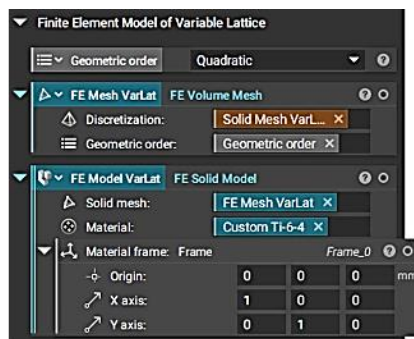
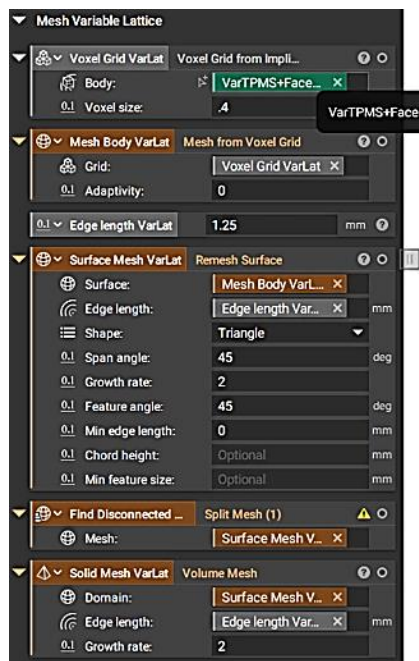
2: 0

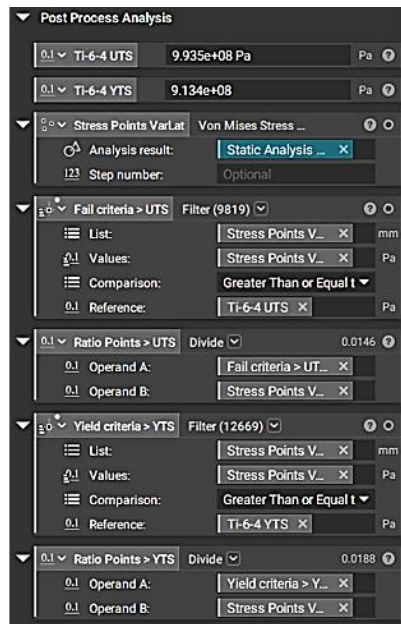
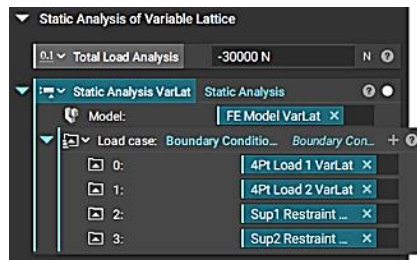
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4: 0

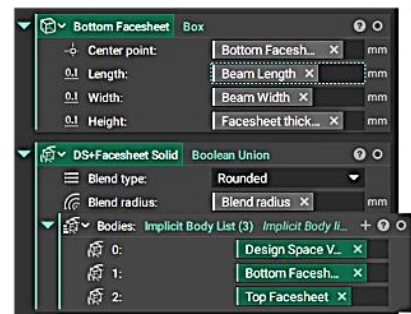


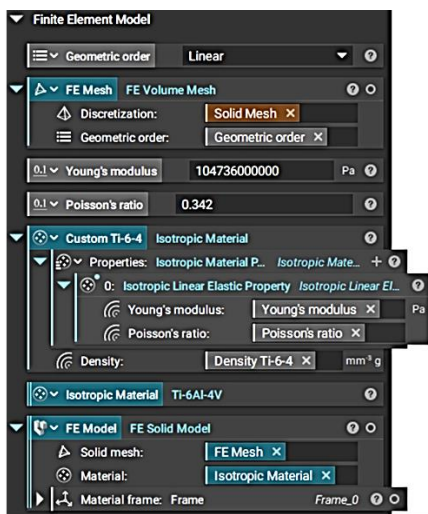
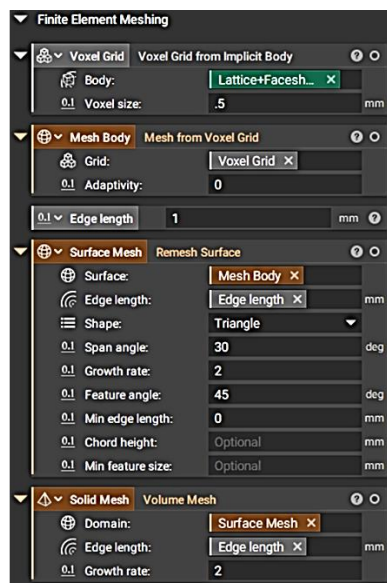
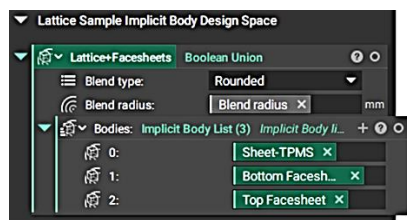
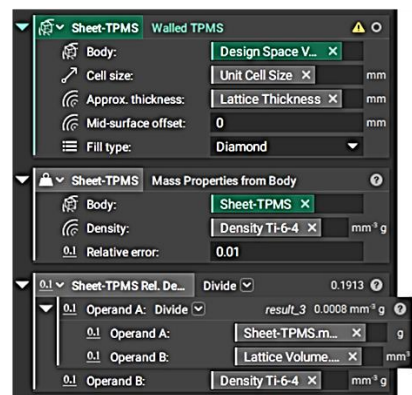
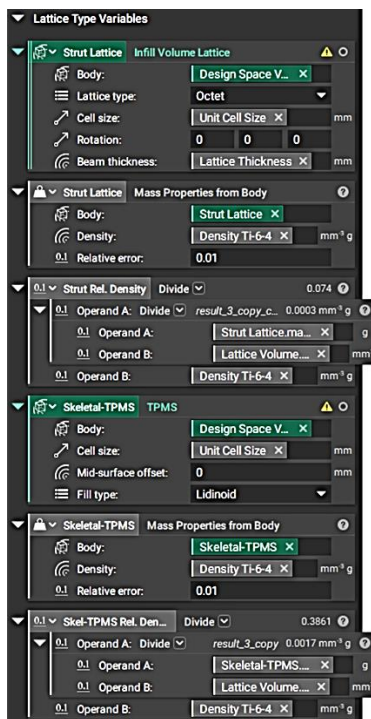
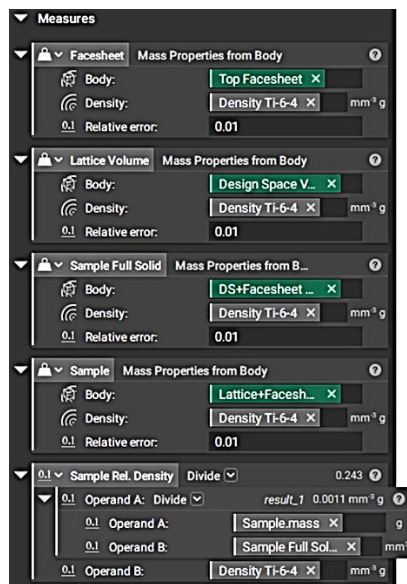






von Mises Stress Minimization Stress Graded Model





Loads and Boundary Locations

Cylinder Radius 2.5 mm

Cylinder Scale Factor 1.1

Span Subtract 142.00 mm

Operand A: Beam Length mm

Operand B: 50 mm

Span Support X Divide 71.00 mm

Operand A: Span mm

Operand B: 2

Cylinder Y Divide 25.00 mm

Operand A: Add result_7 50.00 mm

Operand A: Beam Width mm

Operand B: 2 mm

Cylinder Z Add 15.8564 mm

Operand A: Top Facesheet Z mm

Operand B: Multiply result_14 2.75 mm

Operand A: Cylinder Radius mm

Operand B: Cylinder Scale F. mm

Support-1 P1 Point

X: Span Support X mm

Y: Cylinder Y.negative mm

Z: Cylinder Z.negative mm

Support-1 P2 Point

X: Span Support X mm

Y: Cylinder Y mm

Z: Cylinder Z.negative mm

Span Support 1 Cylinder

Point 1: Support-1 P1 mm

Point 2: Support-1 P2 mm

Radius: Cylinder Radius mm

Support1 Boundary FE Boundary by B...

Mesh: FE Mesh

Entity: Nodes

Body: Span Support 1

Reverse: ☐

Tolerance: 0.001 mm

Support-2 P1 Point

X: Span Support X mm

Y: Cylinder Y.negative mm

Z: Cylinder Z.negative mm

Support-2 P2 Point

X: Span Support X mm

Y: Cylinder Y mm

Z: Cylinder Z.negative mm

Span Support 2 Cylinder

Point 1: Support-2 P1 mm

Point 2: Support-2 P2 mm

Radius: Cylinder Radius mm

Support2 Boundary FE Boundary by B...

Mesh: FE Mesh

Entity: Nodes

Body: Span Support 2

Reverse: ☐

Tolerance: 0.001 mm

Mid Load P1 Point

X: 0 mm

Y: Cylinder Y.negative mm

Z: Cylinder Z mm

Mid Load P2 Point

X: 0 mm

Y: Cylinder Y mm

Z: Cylinder Z mm

Midsan Load Cylinder

Point 1: Mid Load P1 mm

Point 2: Mid Load P2 mm

Radius: Cylinder Radius mm

3pt Load Boundary FE Boundary by Bo...

Mesh: FE Mesh

Entity: Nodes

Body: Midsan Load

Reverse: ☐

Tolerance: 0.001 mm

Load Span Divide 23.6667 mm

Operand A: Span mm

Operand B: Spanwise Load mm

Load1 P1 Point

X: Load Span mm

Y: Cylinder Y.negative mm

Z: Cylinder Z mm

Load1 P2 Point

X: Load Span mm

Y: Cylinder Y mm

Z: Cylinder Z mm

Load Point 1 Cylinder

Point 1: Load1 P1 mm

Point 2: Load1 P2 mm

Radius: Cylinder Radius mm

4pt P1 Boundary FE Boundary by Body

Mesh: FE Mesh

Entity: Nodes

Body: Load Point 1

Reverse: ☐

Tolerance: 0.001 mm

Load2 P1 Point

X: Load Span.negative mm

Y: Cylinder Y.negative mm

Z: Cylinder Z mm

Load2 P2 Point

X: Load Span.negative mm

Y: Cylinder Y mm

Z: Cylinder Z mm

Load Point 2 Cylinder

Point 1: Load2 P1 mm

Point 2: Load2 P2 mm

Radius: Cylinder Radius mm

4pt P2 Boundary FE Boundary by Body

Mesh: FE Mesh

Entity: Nodes

Body: Load Point 2

Reverse: ☐

Tolerance: 0.001 mm

3pt Load Vector Vector (0.00, 0.00, -30.00 N)

X: 0 N

Y: 0 N

Z: Total Load N

4pt Load Vector Vector (0.00, 0.00, -15.00 N)

X: 0 N

Y: 0 N

Z: Total Load N

Point Map 4pt Load Vis

Points: Point List (2)

0: Load Point 1 bo mm

1: Load Point 2 bo mm

Values: Vector List (2)

0: 4pt Load Vector N

1: 4pt Load Vector N

3Pt Applied Load Force

Boundary: 3pt Load Bound mm

Vector: 3pt Load Vector N

Frame: Optional

4Pt Applied Load 1 Force

Boundary: 4pt P1 Boundary mm

Vector: 4pt Load Vector N

Frame: Optional

4Pt Applied Load 2 Force

Boundary: 4pt P2 Boundary mm

Vector: 4pt Load Vector N

Frame: Optional

Allowing Y rotation

Support 1 Restraint Displacement Restr...

Boundary: Support1 Bound mm

Ux: 0 mm

Uy: 0 mm

Uz: 0 mm

Rx: 0 deg

Ry: Optional deg

Rz: 0 deg

Frame: Optional

Support 2 Restraint Displacement Restr...

Boundary: Support2 Bound mm

Ux: 0 mm

Uy: 0 mm

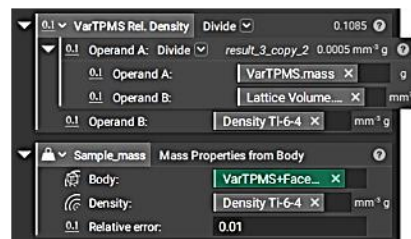
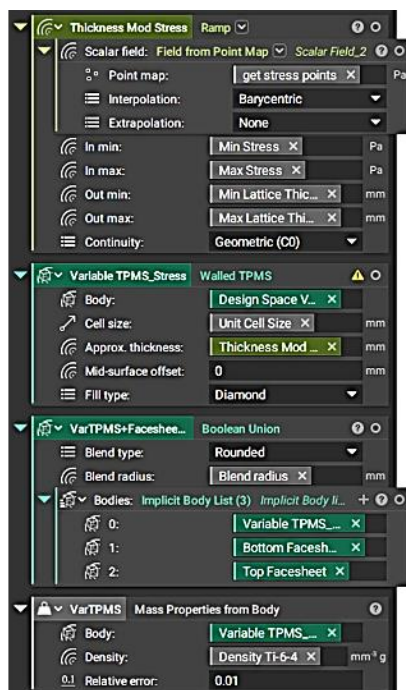
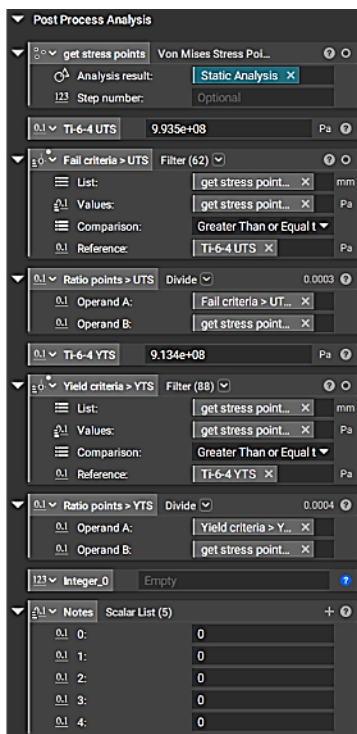
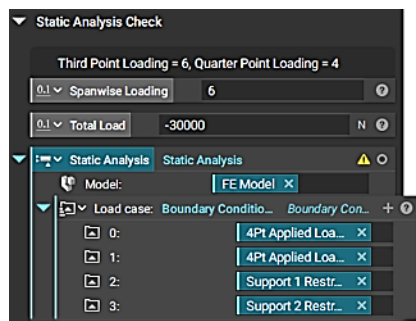
Uz: 0 mm

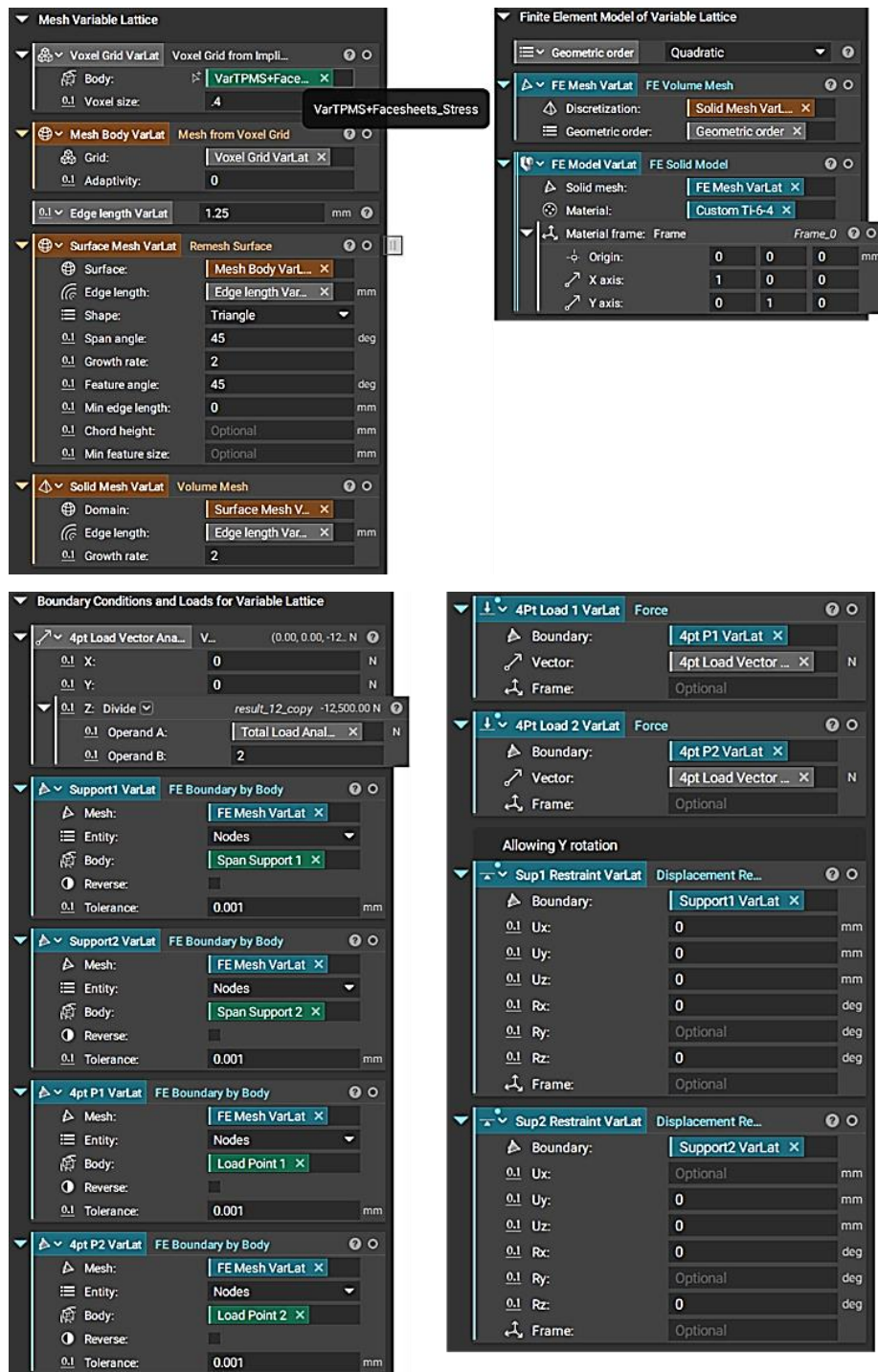
Rx: 0 deg

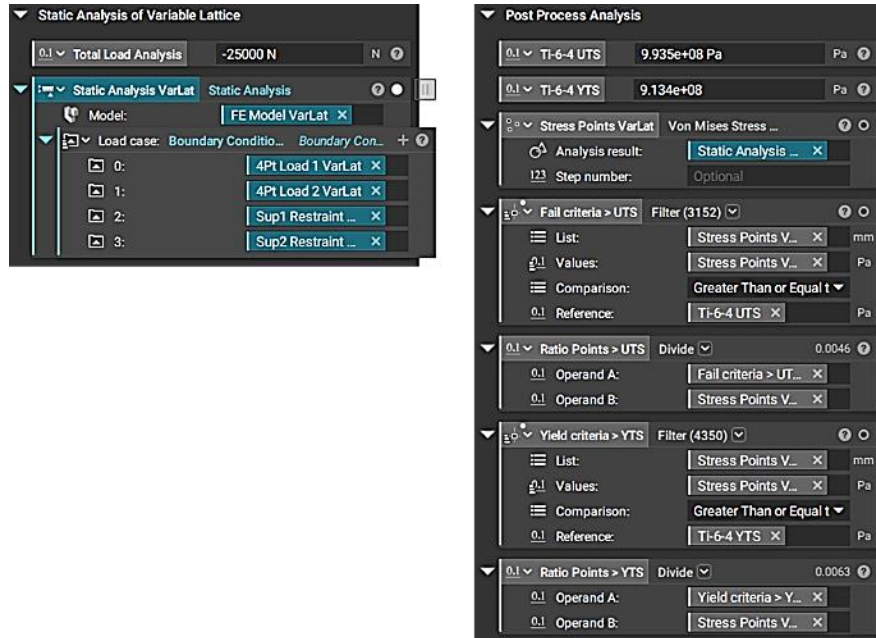
Ry: Optional deg

Rz: 0 deg

Frame: Optional







Beam Modeling Parameters

Specimen number	Grading Method	Lattice Type	Lattice Name	Beam Dimensions				Facesheet Thickness	Unit Cell Size			Blend radius
				Length mm	width mm	Height mm			Length mm	Width mm	Height mm	
1.1	Uniform	TPMS	Gyroid	192	48	25.4	0.8128		24	24	25.4	0.8
1.5	Uniform	TPMS	Primitive	192	48	25.4	0.8128		24	24	25.4	0.8
1.2	Uniform	TPMS	Diamond	192	48	25.4	0.8128		24	24	25.4	0.8
1.4	Uniform	Beam	Octet	192	48	25.4	0.8128		23.6	22.4	25.4	0.8
1.3	Uniform	Beam	Kelvin	192	48	25.4	0.8128		17.2	18.6	17.6	0.8
2.1	Density	TPMS	Gyroid	192	48	25.4	0.8128		24	24	25.4	0.8
2.5	Density	TPMS	Primitive	192	48	25.4	0.8128		24	24	25.4	0.8
2.2	Density	TPMS	Diamond	192	48	25.4	0.8128		24	24	25.4	0.8
2.4	Density	Beam	Octet	192	48	25.4	0.8128		23.6	22.4	25.4	0.8
2.3	Density	Beam	Kelvin	192	48	25.4	0.8128		17.2	18.6	17.6	0.8
3.1	Stress	TPMS	Gyroid	192	48	25.4	0.8128		24	24	25.4	0.8
3.5	Stress	TPMS	Primitive	192	48	25.4	0.8128		24	24	25.4	0.8
3.2	Stress	TPMS	Diamond	192	48	25.4	0.8128		24	24	25.4	0.8
3.4	Stress	Beam	Octet	192	48	25.4	0.8128		23.6	22.4	25.4	0.8
3.3	Stress	Beam	Kelvin	192	48	25.4	0.8128		17.2	18.6	17.6	0.8

Specimen number	Grading Method	Lattice Type	Lattice Name	Lattice		Sim Lattice		Graded Lattice				
				Thickness mm	t/l	Thickness mm	Min Thick mm	Min Value	% min	Max Thick mm	Max Value	% max
1.1	Uniform	TPMS	Gyroid	1.3	0.053	-	-	-	-	-	-	-
1.5	Uniform	TPMS	Primitive	0.75	0.031	-	-	-	-	-	-	-
1.2	Uniform	TPMS	Diamond	1.1	0.045	-	-	-	-	-	-	-
1.4	Uniform	Beam	Octet	2.2	0.092	-	-	-	-	-	-	-
1.3	Uniform	Beam	Kelvin	2.4	0.135	-	-	-	-	-	-	-
2.1	Density	TPMS	Gyroid	variable	-	4	0.6	0.05	-	2.0	0.45	-
2.5	Density	TPMS	Primitive	variable	-	3	0.6	0.40	-	2.0	1.00	-
2.2	Density	TPMS	Diamond	variable	-	4	0.6	0.05	-	2.0	0.60	-
2.4	Density	Beam	Octet	variable	-	4	1.5	0.05	-	3.5	0.55	-
2.3	Density	Beam	Kelvin	variable	-	4	1.5	0.05	-	3.5	0.45	-
3.1	Stress	TPMS	Gyroid	variable	-	2	0.6	2.00E+07	13.8	2	2.00E+08	26.7
3.5	Stress	TPMS	Primitive	variable	-	2	0.6	4.00E+07	35.9	2	4.00E+08	2.7
3.2	Stress	TPMS	Diamond	variable	-	2	0.6	3.00E+07	22.4	2	2.50E+08	13.3
3.4	Stress	Beam	Octet	variable	-	3	1.5	2.41E+07	13.6	3	2.00E+08	27.3
3.3	Stress	Beam	Kelvin	variable	-	3	1.5	3.00E+07	14.3	3	3.00E+08	28.2

Specimen number	Grading Method	Lattice Type	Lattice Name	Lattice				
				Mass g	Volume mm^3	Magics Vol. old	Rel. Den. new	Rel. Den.
1.1	Uniform	TPMS	Gyroid	171.03	38608.1		0.103	0.103
1.5	Uniform	TPMS	Primitive	174.43	39375.5		0.106	0.106
1.2	Uniform	TPMS	Diamond	173.07	39067.7		0.105	0.105
1.4	Uniform	Beam	Octet	173.23	39103.4		0.105	0.105
1.3	Uniform	Beam	Kelvin	174.93	39487.8		0.108	0.108
2.1	Density	TPMS	Gyroid	168.89	38123.1		0.101	0.101
2.5	Density	TPMS	Primitive	173.17	39090.9		0.104	0.104
2.2	Density	TPMS	Diamond	170.32	38448		0.103	0.103
2.4	Density	Beam	Octet	171.29	38665.2		0.103	0.103
2.3	Density	Beam	Kelvin	176.23	39780.9		0.104	0.109
3.1	Stress	TPMS	Gyroid	175.07	39520.2		0.107	0.107
3.5	Stress	TPMS	Primitive	178.08	40198.6		0.109	0.109
3.2	Stress	TPMS	Diamond	176.37	39813.8		0.109	0.109
3.4	Stress	Beam	Octet	180.15	40665.6		0.104	0.111
3.3	Stress	Beam	Kelvin	171.31	38671.4		0.102	0.104

0.104979

Specimen number	Grading Method	Lattice Type	Lattice Name	FEA mesh			FEA 30 kN			
				Voxel Size mm	Edge mm	Span Angle deg	Element Count	Displacement mm	VM > YTS % nodes	VM > UTS % nodes
1.1	Uniform	TPMS	Gyroid	0.3	0.7	30	9.30E+05	1.30	2.51	2.02
1.5	Uniform	TPMS	Primitive	0.3	0.7	30	8.34E+05	1.22	1.05	0.85
1.2	Uniform	TPMS	Diamond	0.3	0.7	30	1.15E+06	1.38	1.89	1.45
1.4	Uniform	Beam	Octet	0.4	1	45	7.58E+05	1.33	2.66	1.97
1.3	Uniform	Beam	Kelvin	0.4	1	30	8.99E+05	2.70	13.70	11.50
2.1	Density	TPMS	Gyroid	0.3	0.7	30	9.40E+05	1.22	1.71	1.44
2.5	Density	TPMS	Primitive	0.3	0.7	30	8.26E+05	1.32	1.94	1.26
2.2	Density	TPMS	Diamond	0.3	0.7	30	1.07E+06	1.28	3.27	2.74
2.4	Density	Beam	Octet	0.3	1	45	8.58E+05	1.97	5.38	4.13
2.3	Density	Beam	Kelvin	0.3	1	45	6.61E+05	1.31	1.88	1.46
3.1	Stress	TPMS	Gyroid	0.3	0.7	30	1.01E+06	1.14	1.35	1.11
3.5	Stress	TPMS	Primitive	0.4	0.7	30	8.43E+05	1.11	0.66	0.50
3.2	Stress	TPMS	Diamond	0.3	1	30	7.24E+05	1.15	1.13	0.89
3.4	Stress	Beam	Octet	0.3	1	45	1.17E+06	1.12	1.29	1.00
3.3	Stress	Beam	Kelvin	0.3	1.25	45	9.92E+05	1.85	4.54	3.40

9.05E+05

Specimen number	Grading Method	Lattice Type	Lattice Name	Build										Specimen Count
				6 success	7 partial	8 failed	9 partial	10 success	11 success	14 success	18 success	19 success		
1.1	Uniform	TPMS	Gyroid	U1	U2		U3	U4	UV	AU			6	
1.5	Uniform	TPMS	Primitive	U1	U2		U3	U4	UV	AU			6	
1.2	Uniform	TPMS	Diamond	U1	U2		U3	U4	UV	AU			6	
1.4	Uniform	Beam	Octet	U1	U2		U3	U4	UV	AU			6	
1.3	Uniform	Beam	Kelvin	U1	U2		U3	U4	UV	AU			6	
2.1	Density	TPMS	Gyroid	D1	D2		D3	D4	DV		AD		6	
2.5	Density	TPMS	Primitive	D1	D2		D3	D4	DV		AD		6	
2.2	Density	TPMS	Diamond	D1	D2		D3	D4	DV		AD		6	
2.4	Density	Beam	Octet	D1	D2		D3	D4	DV		AD		6	
2.3	Density	Beam	Kelvin	D1	D2		D3	D4	DV		AD		6	
3.1	Stress	TPMS	Gyroid	S1			S3	S4	SV			AS	5	
3.5	Stress	TPMS	Primitive	S1	S2		S3	S4	SV			AS	6	
3.2	Stress	TPMS	Diamond	S1	S2		S3	S4	SV			AS	6	
3.4	Stress	Beam	Octet	S1			S3	S4	SV			AS	5	
3.3	Stress	Beam	Kelvin	S1	S2		S3	S4	SV			AS	6	
				14	12	0	14	14	14	5	5	4	82	

Mass/Volume/Density tables:

Specimen ID	Lattice	Grading	Length (mm)	Width (mm)	Height (mm)	STL Mass (g)	STL Volume (ml)	STL Density (g/cc)	STL relative Density	Volume Implicit model (mm^3)	Volume STL model (mm^3)	Volume % diff Implicit vs STL
NA	Gyroid	Uniform	192	48	27.03	171.03	38.8	4.409	0.103	38608.1	38787.764	0.47
NA	Diamond	Uniform	192	48	27.03	173.07	39.2	4.414	0.105	39067.7	39207.655	0.36
NA	Kelvin	Uniform	192	48	27.03	174.93	38.4	4.556	0.108	39487.8	38395.006	-2.77
NA	Octet	Uniform	192	48	27.03	173.23	37.7	4.599	0.105	39103.4	37666.92	-3.67
NA	Primitive	Uniform	192	48	27.03	174.43	39.5	4.412	0.106	39375.5	39531.891	0.40
NA	Gyroid	Density	192	48	27.03	168.89	38.3	4.415	0.101	38123.1	38253.391	0.34
NA	Diamond	Density	192	48	27.03	170.32	38.6	4.409	0.103	38448	38633.842	0.48
NA	Kelvin	Density	192	48	27.03	176.23	37.6	4.681	0.109	39780.9	37646.297	-5.37
NA	Octet	Density	192	48	27.03	171.29	37.6	4.554	0.103	38665.2	37612.043	-2.72
NA	Primitive	Density	192	48	27.03	173.17	39.1	4.425	0.104	39090.9	39133.385	0.11
NA	Gyroid	Stress	192	48	27.03	175.07	39.7	4.413	0.107	39520.2	39673.99	0.39
NA	Diamond	Stress	192	48	27.03	176.37	40.1	4.403	0.109	39813.8	40054.418	0.60
NA	Kelvin	Stress	192	48	27.03	171.31	37.1	4.614	0.104	38671.4	37125.96	-4.00
NA	Octet	Stress	192	48	27.03	180.15	38.7	4.654	0.111	40665.6	38709.434	-4.81
NA	Primitive	Stress	192	48	27.03	178.08	40.3	4.417	0.109	40198.6	40314.63	0.29
Average Model			192	48	27.03	173.84	38.7	4.492	0.106	39241.3	38716.4	-1.33

Specimen ID	Lattice	Grading	#	Build	Length (mm)	Width (mm)	Height (mm)	Mass (g)	Volume (ml)	Density (g/cc)
1	Kelvin	Uniform	1	6	190.52	47.61	26.95	147.32	35	4.209
2	Octet	Uniform	1	6	190.77	47.97	26.94	145.44	35	4.155
3	Gyroid	Uniform	1	6	190.7	47.95	26.83	155.59	35	4.445
4	Diamond	Uniform	1	6	190.73	47.63	26.93	152.69	35	4.363
5	Primitive	Uniform	1	6	190.84	47.49	27.07	150.65	35	4.304
6	Kelvin	Density	1	6	190.88	48.27	26.87	148.32	35	4.238
7	Octet	Density	1	6	190.66	48.13	26.96	147.92	35	4.226
8	Gyroid	Density	1	6	190.66	48.13	26.96	156.06	35	4.459
9	Diamond	Density	1	6	190.79	48.13	26.99	155.73	35	4.449
10	Primitive	Density	1	6	190.82	48.12	27.26	154.43	40	3.861
11	Kelvin	Stress	1	6	190.66	48.26	26.9	145.86	35	4.167
12	Octet	Stress	1	6	190.67	48.51	27.01	152.31	35	4.352
13	Gyroid	Stress	1	6	190.71	48.31	27.06	165	40	4.125
14	Diamond	Stress	1	6	190.79	48.23	27.03	164.21	35	4.692
15	Primitive	Stress	1	6	191.02	48.28	27.31	157.67	40	3.942

Specimen ID	Lattice	Grading	#	Build	Length (mm)	Width (mm)	Height (mm)	Mass (g)	Volume (ml)	Density (g/cc)
16	Kelvin	Uniform	2	7	190.62	47.28	26.88	149.25	35	4.264
17	Octet	Uniform	2	7	190.59	47.93	26.87	147.14	35	4.204
18	Gyroid	Uniform	2	7	190.94	47.93	26.93	152.64	35	4.361
19	Diamond	Uniform	2	7	190.82	48.23	27.05	159.05	35	4.544
20	Primitive	Uniform	2	7	190.73	47.37	26.99	151.09	40	3.777
21	Kelvin	Density	2	7	190.63	48.28	26.94	150.5	30	5.017
22	Octet	Density	2	7	190.65	48.35	27.01	149.73	35	4.278
23	Gyroid	Density	2	7	190.91	48.43	27.07	152.15	35	4.347
24	Diamond	Density	2	7	190.58	47.36	26.99	155.4	35	4.440
25	Primitive	Density	2	7	190.93	48.33	27.17	154.75	35	4.421

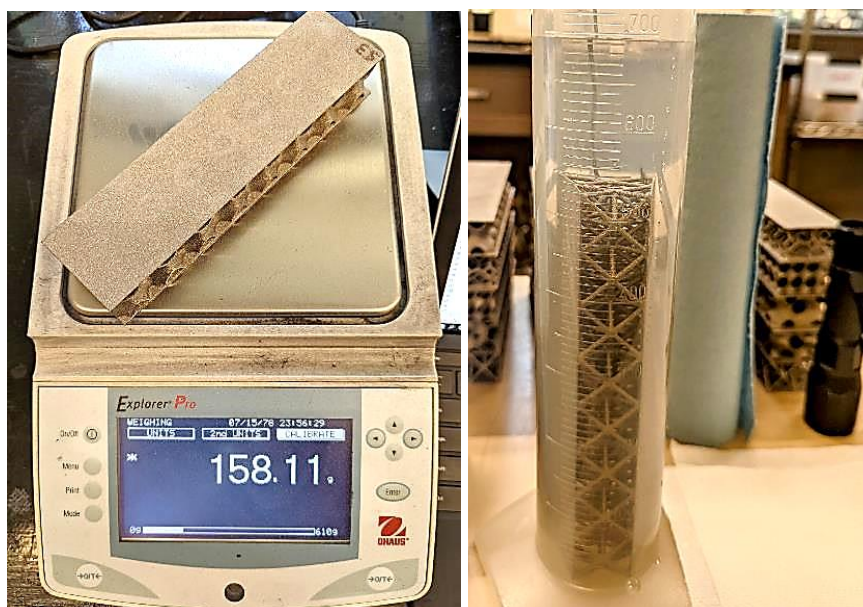
Specimen ID	Lattice	Grading	#	Build	Length (mm)	Width (mm)	Height (mm)	Mass (g)	Volume (ml)	Density (g/cc)
26	Kelvin	Uniform	3	9	190.56	47.29	26.9	147.73	30	4.924
27	Octet	Uniform	3	9	190.73	47.8	27.09	146.4	35	4.183
28	Gyroid	Uniform	3	9	190.83	47.88	27.01	151.85	35	4.339
29	Diamond	Uniform	3	9	190.74	47.12	27.05	147.1	35	4.203
30	Primitive	Uniform	3	9	190.9	47.26	27.67	148.98	30	4.966
31	Kelvin	Density	3	9	190.84	48.09	26.89	149.39	35	4.268
32	Octet	Density	3	9	190.83	48.06	26.93	148.35	35	4.239
33	Gyroid	Density	3	9	190.77	48.1	27.02	151.41	35	4.326
34	Diamond	Density	3	9	190.84	48.07	27.06	157.69	35	4.505
35	Primitive	Density	3	9	190.92	48.13	27.03	153.38	35	4.382
36	Kelvin	Stress	3	9	190.82	48.32	26.96	146.82	35	4.195
37	Octet	Stress	3	9	190.82	48.23	27.02	153.4	35	4.383
38	Gyroid	Stress	3	9	190.73	48.22	26.92	158.14	35	4.518
39	Diamond	Stress	3	9	190.8	48.2	27.09	165.68	40	4.142
40	Primitive	Stress	3	9	190.81	48.4	27.08	158.11	35	4.517

Specimen ID	Lattice	Grading	#	Build	Length (mm)	Width (mm)	Height (mm)	Mass (g)	Volume (ml)	Density (g/cc)
41	Kelvin	Uniform	4	10	190.63	47.35	26.9	148.1	35	4.231
42	Octet	Uniform	4	10	190.92	47.76	26.9	146.4	35	4.183
43	Gyroid	Uniform	4	10	191.41	47.83	26.92	150.44	35	4.298
44	Diamond	Uniform	4	10	190.84	47.38	27.02	149.35	35	4.267
45	Primitive	Uniform	4	10	190.9	47.32	27.01	150.38	35	4.297
46	Kelvin	Density	4	10	190.81	48.12	26.93	148.59	35	4.245
47	Octet	Density	4	10	191.41	48.13	26.94	147.89	35	4.225
48	Gyroid	Density	4	10	191.14	48.27	26.91	150.04	35	4.287
49	Diamond	Density	4	10	190.87	48.06	27.05	156.54	35	4.473
50	Primitive	Density	4	10	190.97	48.18	27.07	152.87	40	3.822
51	Kelvin	Stress	4	10	190.75	48.05	26.9	146.29	30	4.876
52	Octet	Stress	4	10	190.87	48.21	26.93	153.16	35	4.376
53	Gyroid	Stress	4	10	190.89	48.27	26.92	157.78	35	4.508
54	Diamond	Stress	4	10	190.93	48.2	27.07	165.08	40	4.127
55	Primitive	Stress	4	10	191.22	48.16	27.07	157.48	40	3.937

Specimen ID	Lattice	Grading	#	Build	Length (mm)	Width (mm)	Height (mm)	Mass (g)	Volume (ml)	Density (g/cc)
56	Kelvin	Uniform	V	11	190.82	47.59	26.93	147.86	35	4.225
57	Octet	Uniform	V	11	190.63	47.55	26.88	146.02	30	4.867
58	Gyroid	Uniform	V	11	190.61	47.68	26.96	147.79	35	4.223
59	Diamond	Uniform	V	11	190.62	47.68	27.24	152.02	35	4.343
60	Primitive	Uniform	V	11	190.84	47.62	26.96	150.81	35	4.309
61	Kelvin	Density	V	11	190.88	47.57	26.88	146.38	35	4.182
62	Octet	Density	V	11	190.76	47.65	27.18	147.18	35	4.205
63	Gyroid	Density	V	11	190.82	47.6	26.92	147.49	35	4.214
64	Diamond	Density	V	11	190.76	47.66	26.93	153.41	35	4.383
65	Primitive	Density	V	11	190.84	47.6	27.07	150.77	35	4.308
66	Kelvin	Stress	V	11	190.73	47.6	26.89	144.06	35	4.116
67	Octet	Stress	V	11	191.13	47.51	27.18	150.83	35	4.309
68	Gyroid	Stress	V	11	190.73	47.62	26.89	154.53	35	4.415
69	Diamond	Stress	V	11	190.76	47.59	27.05	161.43	35	4.612
70	Primitive	Stress	V	11	191.07	47.67	27.19	154.77	40	3.869

Specimen ID	Lattice	Grading	#	Build	Length (mm)	Width (mm)	Height (mm)	Mass (g)	Volume (ml)	Density (g/cc)
71	Kelvin	Uniform	A	14	191.11	47.71	26.9	145.31	35	4.152
72	Octet	Uniform	A	14	190.63	47.52	26.87	143.81	35	4.109
73	Gyroid	Uniform	A	14	190.86	47.55	27.02	150.25	35	4.293
74	Diamond	Uniform	A	14	190.81	47.66	27.07	148.55	35	4.244
75	Primitive	Uniform	A	14	191.09	47.52	27.16	146.9	35	4.197
76	Kelvin	Density	A	18	190.78	47.29	26.91	145.59	35	4.160
77	Octet	Density	A	18	190.76	47.18	26.95	145.93	35	4.169
78	Gyroid	Density	A	18	190.78	47.32	26.92	146.83	30	4.894
79	Diamond	Density	A	18	190.88	47.34	26.96	152.53	35	4.358
80	Primitive	Density	A	18	191.08	47.26	27.1	148.66	35	4.247
81	Kelvin	Stress	A	19	190.77	47.28	26.88	142.96	35	4.085
82	Octet	Stress	A	19	190.88	47.53	26.92	150.71	35	4.306
83	Gyroid	Stress	A	19	190.65	47.16	27.26	154.27	35	4.408
84	Diamond	Stress	A	19	190.81	47.44	27	160.64	40	4.016
85	Primitive	Stress	A	19	190.91	47.23	27.01	152.14	35	4.347
Average Produced					190.82	47.81	27.00	151.48	35.24	4.311
CTE calc					190.88	47.72	26.87			

Mass Volume Measurement Setups



Finite element analysis data tables

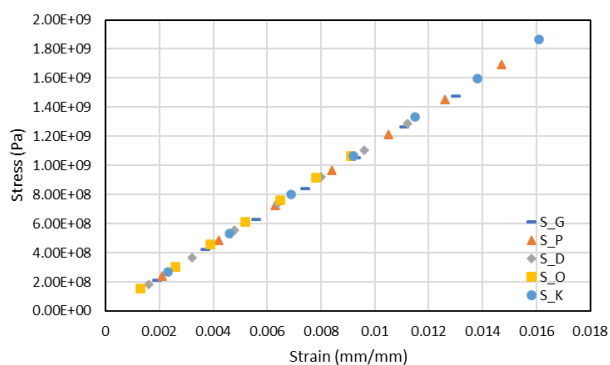
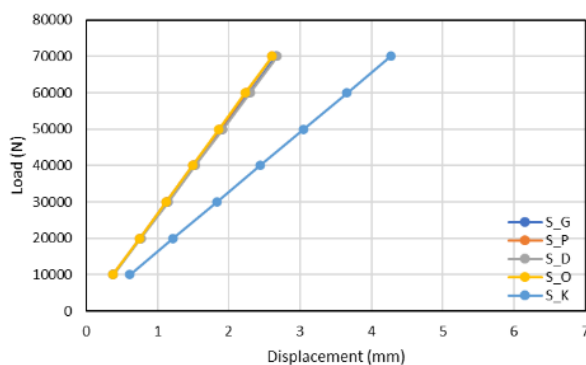
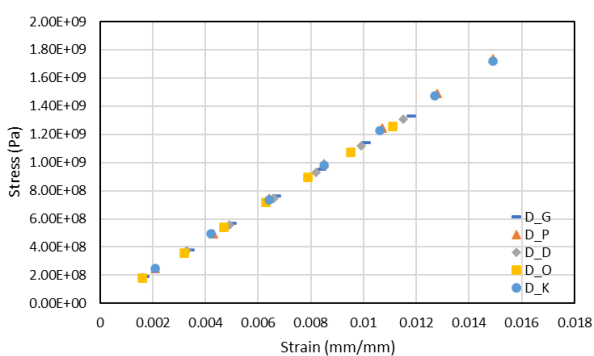
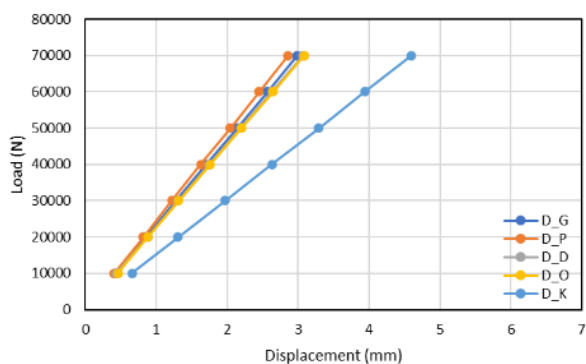
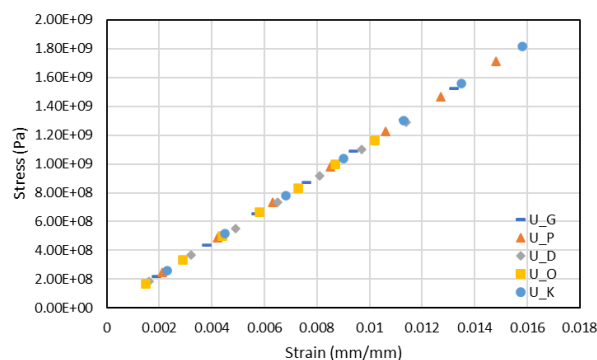
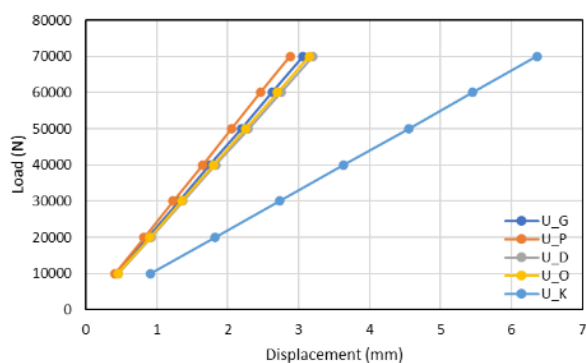
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	Span Angl	30					Span Angl	30					Span Angl	45			
	Elements	448754					Elements	397265					Elements	372174			
Load	Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS
0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0
10000	0.44	2.18E+08	0.0019	0.16	0.12	0.41	2.45E+08	0.0021	0.16	0.12	0.46	1.84E+08	0.0016	0.02	0.01		
20000	0.87	4.36E+08	0.0038	1.14	0.89	0.82	4.90E+08	0.0042	0.69	0.59	0.92	3.68E+08	0.0032	0.5	0.35		
30000	1.31	6.53E+08	0.0057	2.96	2.45	1.23	7.35E+08	0.0063	1.61	1.34	1.37	5.52E+08	0.0049	2.04	1.58		
40000	1.75	8.71E+08	0.0076	6.6	5	1.65	9.80E+08	0.0085	4.88	3.15	1.83	7.35E+08	0.0065	5.4	3.94		
50000	2.19	1.09E+09	0.0094	16.15	11.96	2.05	1.22E+09	0.0106	11.37	8.96	2.29	9.19E+08	0.0081	15.74	10.34		
60000	2.62	1.31E+09	0.0113	24.9	20.89	2.46	1.47E+09	0.0127	17.86	14.71	2.75	1.10E+09	0.0097	24.36	20.46		
70000	3.06	1.52E+09	0.0132	31.74	28.06	2.88	1.71E+09	0.0148	23.72	20.5	3.2	1.29E+09	0.0114	31.83	27.7		

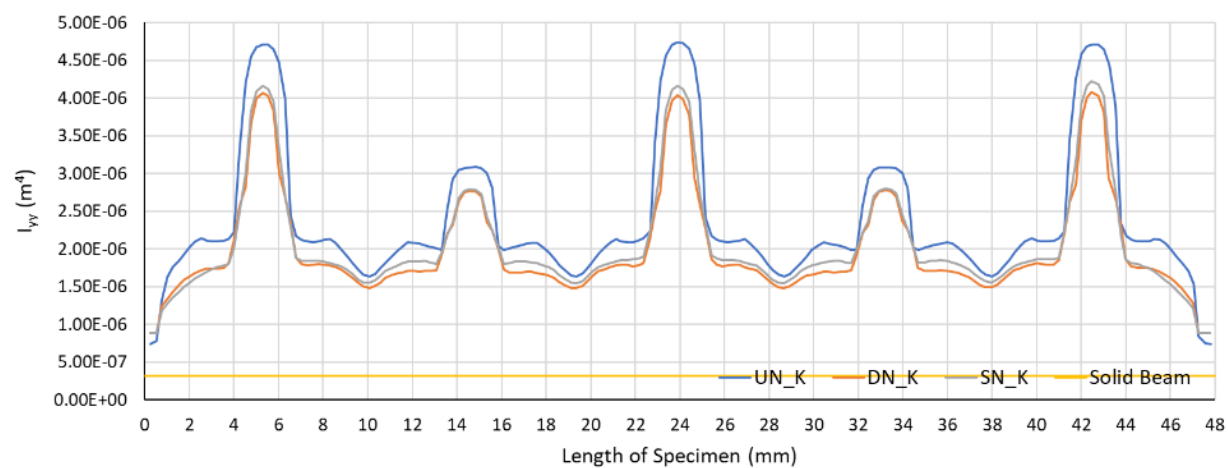
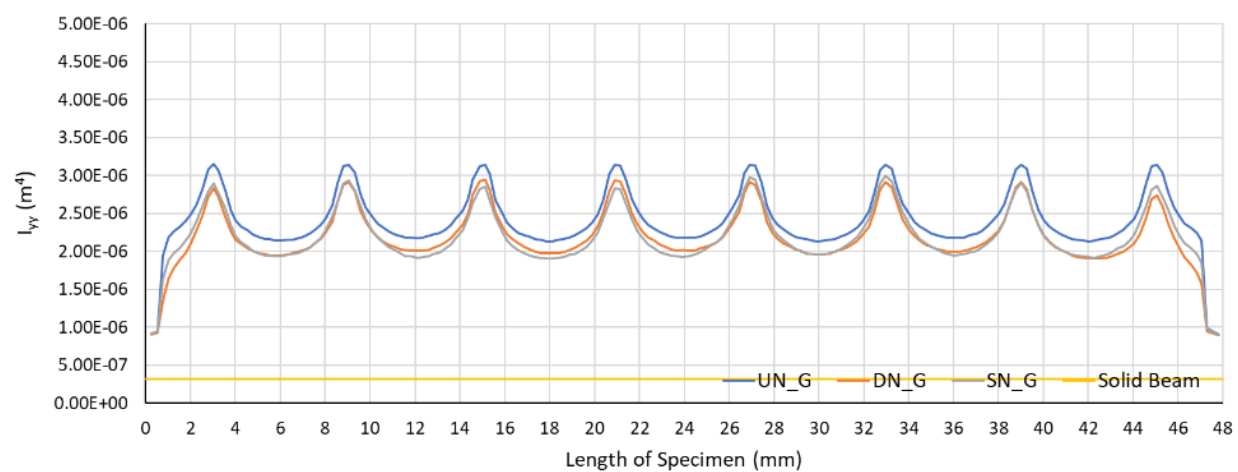
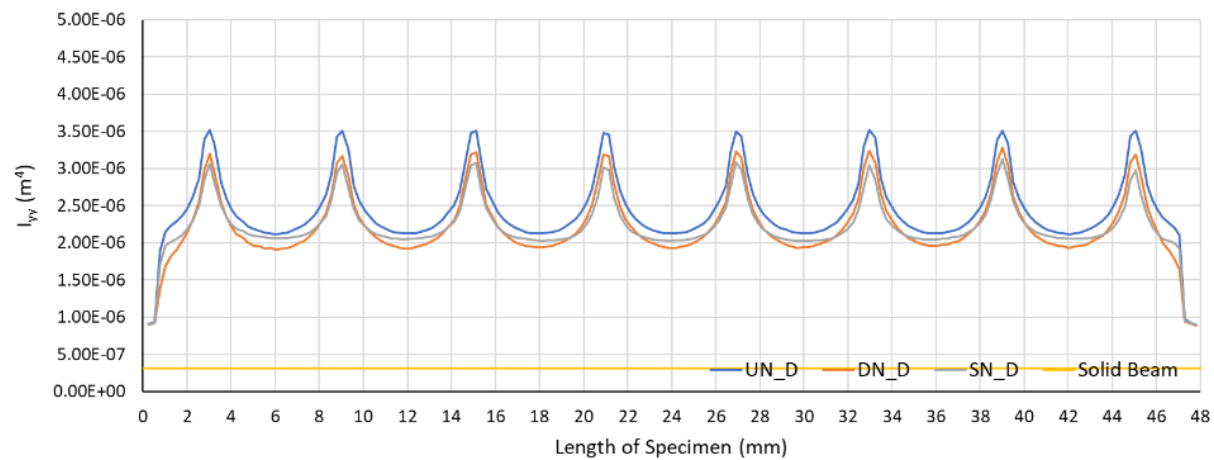
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	Edge Leng	2					Edge Leng	1.25					Edge Leng	1.25			
	Span Angl	60					Span Angl	45					Span Angl	30			
	Elements	425126					Elements	404105					Elements	419407			
Load	Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS
0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0
10000	0.45	1.66E+08	0.0015	0	0	0.91	2.60E+08	0.0023	0.27	0.17	0.43	1.90E+08	0.0017	0.3	0.25		
20000	0.9	3.32E+08	0.0029	0.32	0.19	1.82	5.20E+08	0.0045	5.1	3.9	0.85	3.81E+08	0.0034	1.35	1.1		
30000	1.35	4.98E+08	0.0044	2.28	1.55	2.73	7.79E+08	0.0068	13.93	11.67	1.28	5.71E+08	0.005	3.27	2.74		
40000	1.8	6.63E+08	0.0058	6.05	4.51	3.63	1.04E+09	0.009	24.15	20.63	1.71	7.61E+08	0.0067	6.86	5.22		
50000	2.25	8.29E+08	0.0073	14.59	11.21	4.55	1.30E+09	0.0113	35.01	31.34	2.14	9.51E+08	0.0084	16.72	12.29		
60000	2.7	9.95E+08	0.0087	22.23	18.84	5.45	1.56E+09	0.0135	41.99	38.92	2.57	1.14E+09	0.0101	25.86	21.68		
70000	3.15	1.16E+09	0.0102	28.37	24.93	6.36	1.82E+09	0.0158	47.13	44.42	2.99	1.33E+09	0.0118	32.79	29.09		

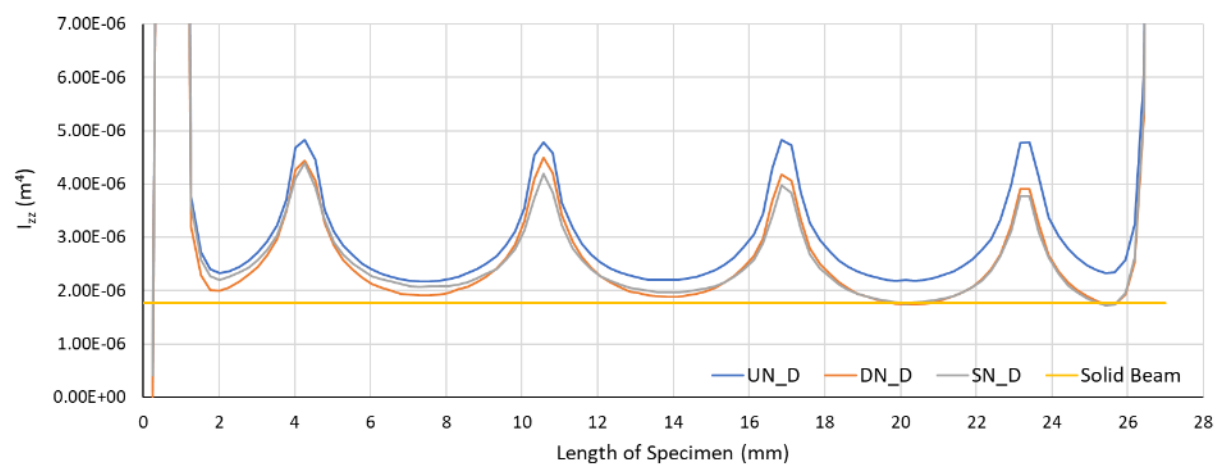
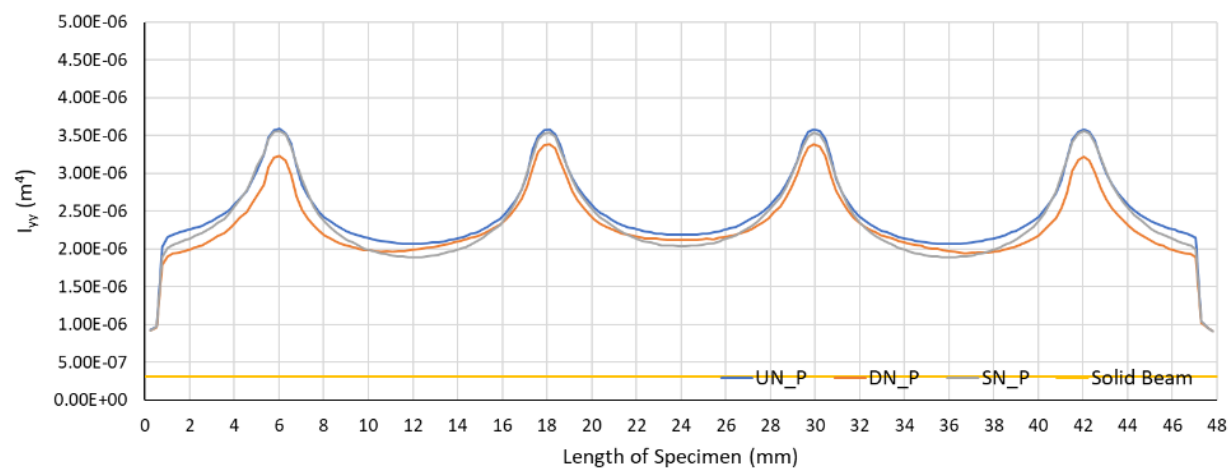
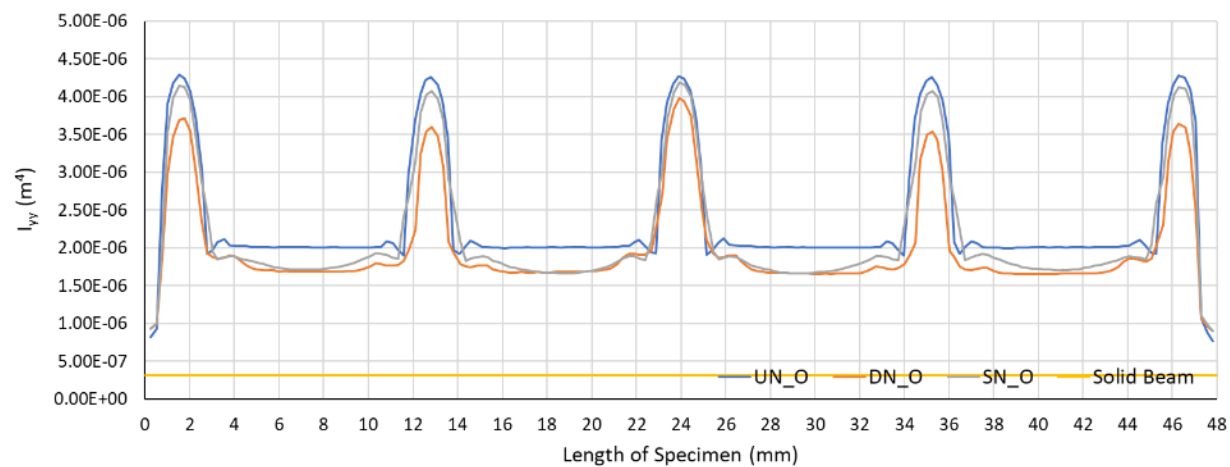
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	Span Angl	30					Span Angl	45					Span Angl	60			
	Elements	420410					Elements	347887					Elements	383473			
Load	Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS
0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0
10000	0.41	2.48E+08	0.0021	0.08	0.06	0.44	1.86E+08	0.0016	0.04	0.02	0.46	1.79E+08	0.0016	0	0		
20000	0.82	4.97E+08	0.0043	0.75	0.61	0.88	3.73E+08	0.0033	0.47	0.34	0.88	3.58E+08	0.0032	0.24	0.14		
30000	1.22	7.45E+08	0.0064	1.71	1.44	1.31	5.59E+08	0.0049	1.88	1.46	1.32	5.38E+08	0.0047	1.94	1.26		
40000	1.63	9.93E+08	0.0085	4.98	3.24	1.75	7.45E+08	0.0066	4.75	3.58	1.76	7.17E+08	0.0063	6.52	4.72		
50000	2.04	1.24E+09	0.0107	12.02	9.32	2.19	9.32E+08	0.0082	14.7	9.1	2.2	8.96E+08	0.0079	12.8	10.37		
60000	2.45	1.49E+09	0.0128	19.01	15.6	2.63	1.12E+09	0.0099	23.85	19.89	2.65	1.08E+09	0.0095	18.45	15.8		
70000	2.86	1.74E+09	0.0149	25.28	21.87	3.07	1.30E+09	0.0115	31.58	27.33	3.09	1.25E+09	0.0111	23.54	20.73		

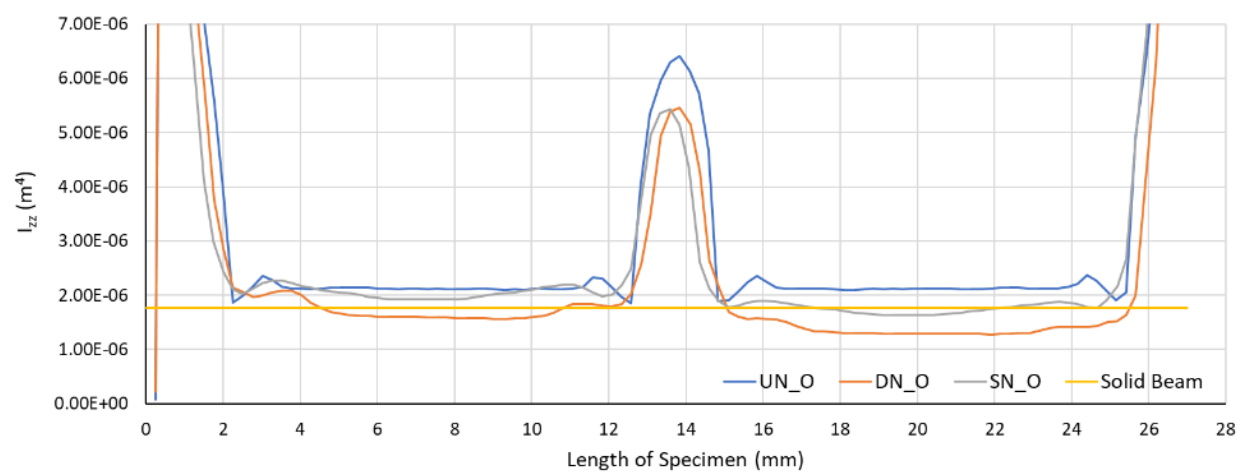
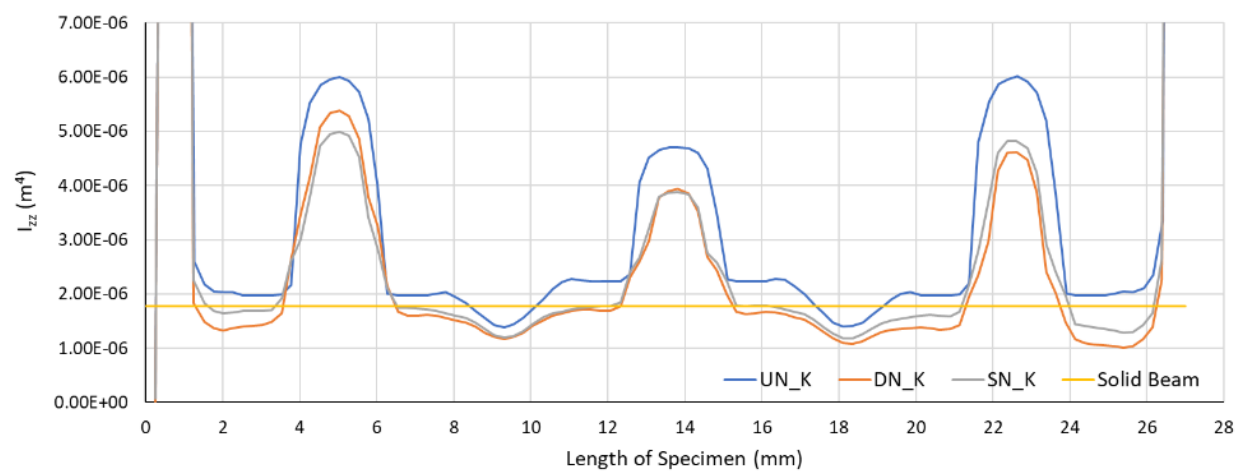
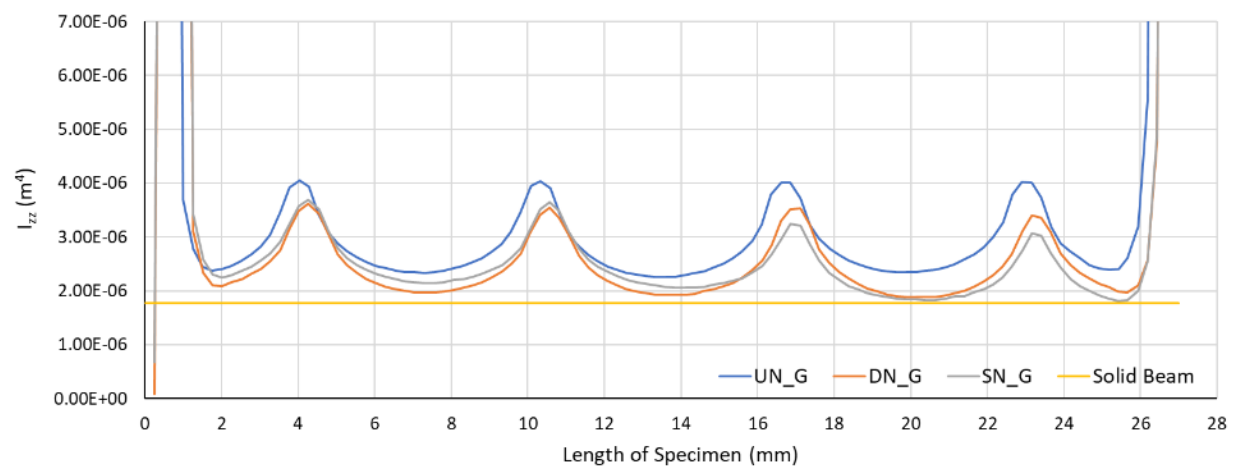
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	Edge Leng	2					Edge Leng	1.25					Edge Leng	1.25			
	Span Angl	60					Span Angl	30					Span Angl	30			
	Elements	281351					Elements	438457					Elements	414926			
Load	Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS		Disp	Stress	Strain	VM > YTS	VM > UTS
0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0
10000	0.66	2.46E+08	0.0021	0.19	0.16	0.38	2.11E+08	0.0019	0.04	0.03	0.37	2.42E+08	0.0021	0.04	0.02		
20000	1.31	4.92E+08	0.0042	1.37	1.02	0.76	4.21E+08	0.0037	0.59	0.45	0.75	4.84E+08	0.0042	0.38	0.31		
30000	1.97	7.37E+08	0.0064	5.38	4.13	1.14	6.32E+08	0.0056	1.75	1.43	1.12	7.27E+08	0.0063	0.96	0.79		
40000	2.63	9.83E+08	0.0085	12.12	9.66	1.52	8.42E+08	0.0074	3.87	2.91	1.49	9.69E+08	0.0084	3.01	1.89		
50000	3.29	1.23E+09	0.0106	21.71	17.85	1.9	1.05E+09	0.0093	10.72	7.58	1.87	1.21E+09	0.0105	8.14	6.04		
60000	3.94	1.47E+09	0.0127	29.31	25.9	2.28	1.26E+09	0.0111	18.5	14.81	2.24	1.45E+09	0.0126	13.6	11.02		
70000	4.6	1.72E+09	0.0149	35.45	32.11	2.66	1.47E+09	0.013	25.09	21.53	2.61	1.70E+09	0.0147	18.86	15.89		

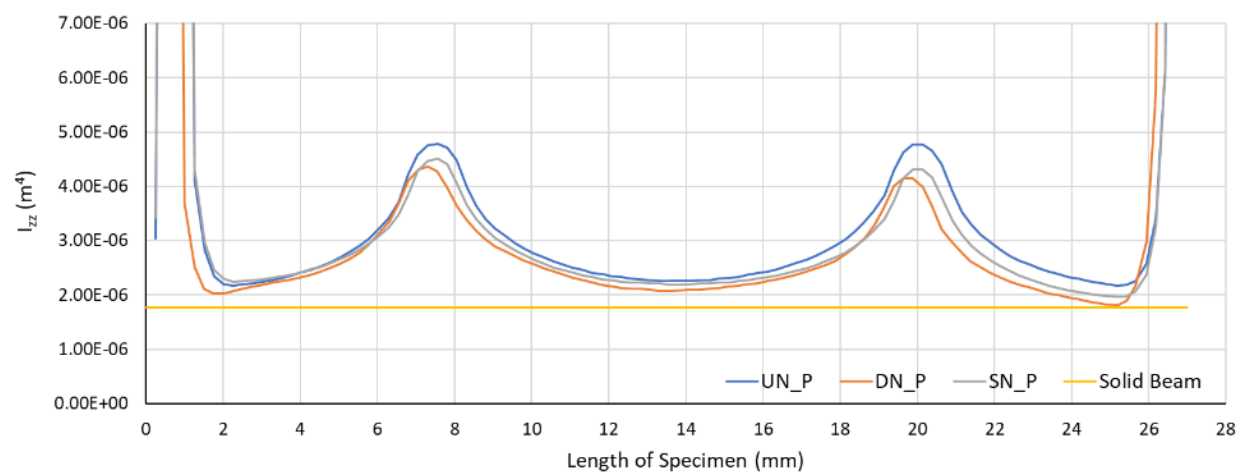
	S_D					S_O					S_K						
	Voxel size	0.4				Voxel size	0.4				Voxel size	0.4					
	Edge Leng	1.25				Edge Leng	2				Edge Leng	2					
	Span Angl	45				Span Angl	60				Span Angl	60					
	Elements	356035				Elements	463241				Elements	361242					
Load	Disp	Stress	Strain	VM > YTS	VM > UTS	Disp	Stress	Strain	VM > YTS	VM > UTS	Disp	Stress	Strain	VM > YTS	VM > UTS		
	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	0	
10000	0.38	1.84E+08	0.0016	0.01	0	0.37	1.52E+08	0.0013	0	0	0.61	2.66E+08	0.0023	0.11	0.09		
20000	0.77	3.68E+08	0.0032	0.25	0.16	0.75	3.05E+08	0.0026	0.06	0.03	1.22	5.32E+08	0.0046	1.02	0.75		
30000	1.15	5.52E+08	0.0048	1.11	0.87	1.12	4.57E+08	0.0039	0.67	0.46	1.83	7.99E+08	0.0069	4.17	3.17		
40000	1.53	7.37E+08	0.0064	2.33	1.86	1.49	6.10E+08	0.0052	2.28	1.54	2.44	1.06E+09	0.0092	10.16	7.75		
50000	1.92	9.21E+08	0.008	8.68	5.79	1.86	7.62E+08	0.0065	5.91	4.42	3.05	1.33E+09	0.0115	18.44	15.09		
60000	2.3	1.10E+09	0.0096	16.39	12.88	2.23	9.15E+08	0.0078	10.31	8.15	3.66	1.60E+09	0.0138	26	22.44		
70000	2.68	1.29E+09	0.0112	23.74	19.56	2.61	1.07E+09	0.0091	14.88	12.21	4.27	1.86E+09	0.0161	32.98	29.15		



Moment of Inertia Plots for I_{yy} and I_{zz} 







APPENDIX C

Arcam A2X Build Theme Process Parameters

A2X Preheat 50um (Preheat each layer)			
Top Level	Focus Offset	70	mA
	Heating Focus Offset	130	mA
	Maximum Heat time	100	s
	Max Adaptive Iterations	20	
Square	Size	144	mm
	offset to part	4	mm
	Max Current	48	mA
Preheating 1 "Jump Safe"	Beam: Max Beam current	30	mA
	Beam: Min Beam current	30	mA
	Beam: Beam speed	10000	mm/s
	Beam: Number of repetitions	2	
	Beam: Average current	0	mA
	Hatch: Line order	15	
	Hatch: Line Offset	1.2	mm
	Hatch: Hatch depth	0.09	mm
Preheating 2 "Melt Safe" <i>preheat the powder circumventing the parts with an additional preheat sequence. If preheat 1 is disabled the entire square will be preheated.</i>	Beam: Max Beam current	38	mA
	Beam: Min Beam current	38	mA
	Beam: Beam speed	13000	mm/s
	Beam: Number of repetitions	3	
	Beam: Average current	12.6	mA
	Hatch: Line order	15	
	Hatch: Line Offset	1.2	mm
	Hatch: Hatch depth	0.09	mm
Wafer Support			
Top Level	Minimum distance	0.9	mm
	Repetitions	2	
	Speed in square	140000	mm/s
	Current	5.5	mA
	Speed in square	1600	mm/s
	Focus offset	4	mA
Melt			
Power Analyse	Enabled	TRUE	
	Surface temp	650	C
	MaxDepthFor CurrAdd	5000	mm
	ScaleCurrAdd	FALSE	
Contours	Number of contours	3	
	Block Offset	TRUE	
Contours (Outer)	Beam Speed	440	mm/s
	Beam Current	5	mA
	Beam Max Current	5	mA
	Beam Focus Offset	6	mA
	Offset	0.29	
	Multispot Speed in Square	140000	mm/s
Contours (Inner)	Beam Speed	850	mm/s
	Beam Current	10.2	mA
	Beam Max Current	10.2	mA
	Beam Focus Offset	6	mA
	Offset	0.17	mm

	Multispot Speed in Square	100000	mm/s
Hatch	Enabled	TRUE	
Hatch Square	UseRotatingHatch	TRUE	
Hatch Beam	Max current	20	mA
	Focus offset	25	mA
	Speed Function	45	
Hatch Hatch	Line offset	0.2	mm
	Hatch depth	0.05	mm
Hatch Thickness	Speed Factor	1.5	
Hatch Turning Point	Enabled	TRUE	
	Pre exponential factor	0.9	
Hatch Current Compensation	Scan length reference	45	mm
	RefCurrent	12	mA
	MinCurrent	3.5	mA
Hatch Line offset compensation	Max line offset	0.2	mm
210 Plate-A2X_Temperature (Initial Build Preheat)			
Sweep	Max Beam Current	48	mA
	Min Beam Current	48	mA
	Speed	35000	mm/s
Size	Width	195	mm
	Height	195	mm
Beam	Focus Offset	80	
Time out	heat start plate timeout	65	min
Clean plate	Enabled	FALSE	
Outgassing	Heat shield outgassing time	15	min
Time Controlled Heating Step	Beam Current	48	
	Time	30	
	Cooling Factor	1	

Build Record

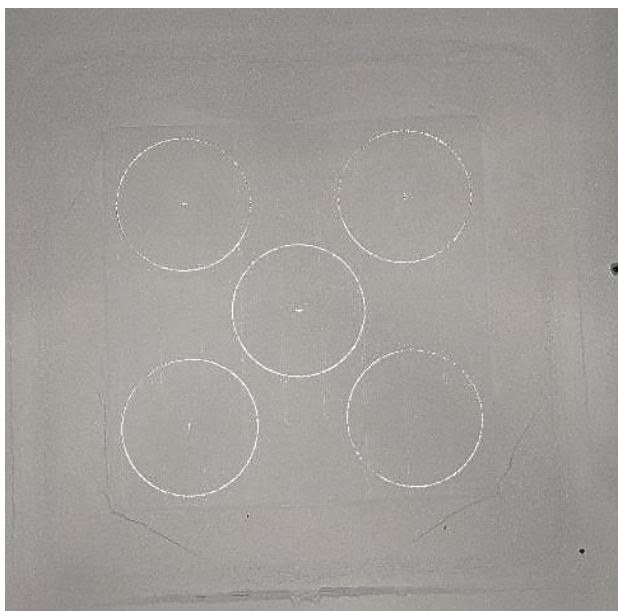
Build #	Result	Config	Feedstock	Date
0	smoked	Line Build	waste L-PBF	1/17/21
1	success	Line Build	waste L-PBF	2/27/21
2	failed	Lattice Mfg v1	waste L-PBF	5/1/21
3	failed	Lattice Mfg v1	waste L-PBF	6/5/21
4	partial	Lattice Mfg v1	waste L-PBF	6/12/21
5	success	Lattice Mfg v2	waste L-PBF	6/26/21
6	success	Lattice Beam v3	waste L-PBF	10/16/21
7	partial	Lattice Beam v3	waste L-PBF	1/4/22
8	failed	Lattice Beam v4	waste L-PBF	1/18/22
9	partial	Lattice Beam v3	waste L-PBF	2/1/22
10	success	Lattice Beam v3	waste L-PBF	2/15/22
11	success	Lattice Vertical	waste L-PBF	3/22/22
12	success	Coupon	waste L-PBF	7/19/22
13	success	Coupon	waste L-PBF	7/21/22
14	success	Beam Uniform	virgin EB-PBF	8/23/22

15	failed	Beam Density	virgin EB-PBF	8/24/22
16	failed	Beam Density	virgin EB-PBF	8/26/22
17	failed	Beam Density	virgin EB-PBF	8/29/22
18	success	Beam Density	virgin EB-PBF	8/30/22
19	success	Beam Stress	virgin EB-PBF	9/1/22

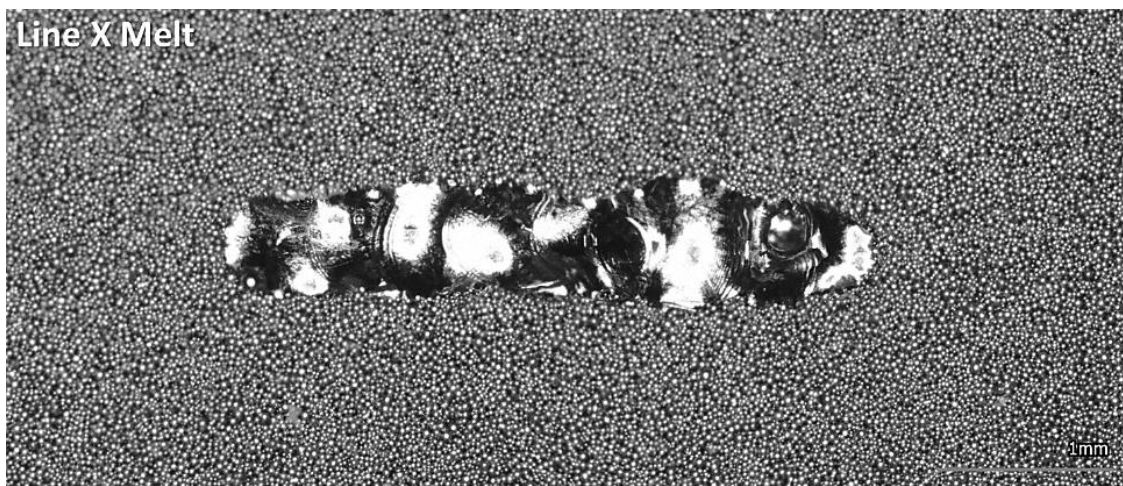
Build 0 Line Build 1/17/21 (smoked)



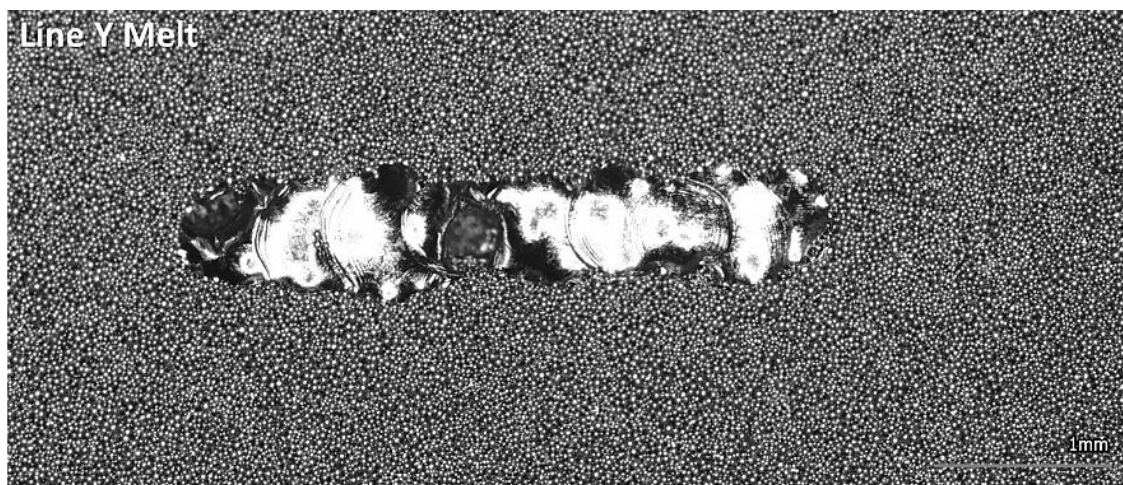
Build 1 Line Build 2/27/21 (success)

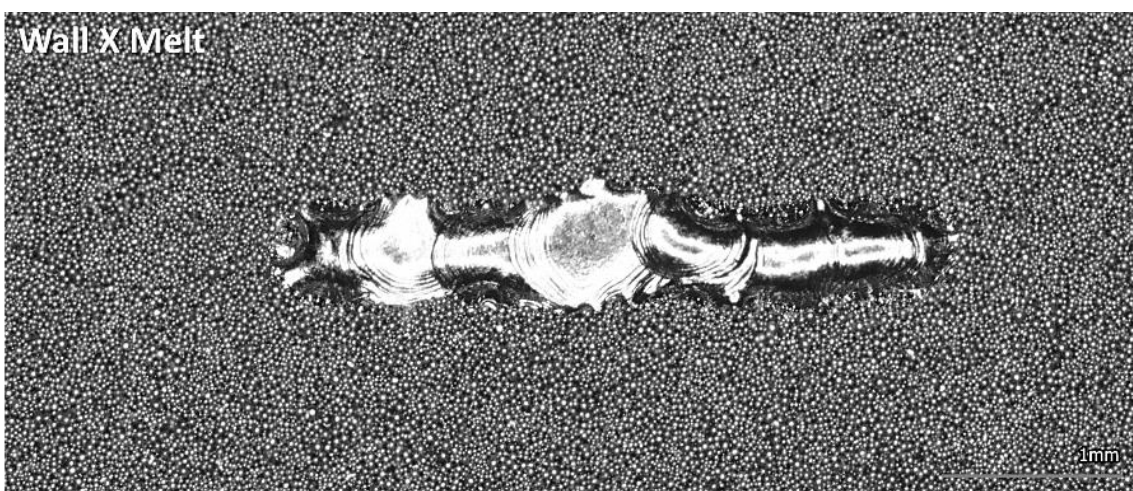
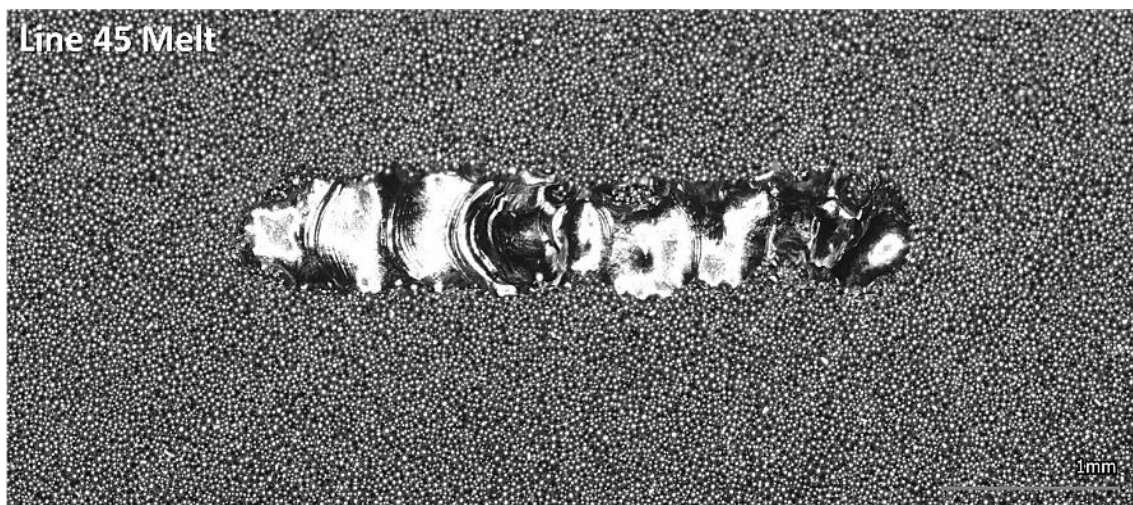


Line X Melt



Line Y Melt



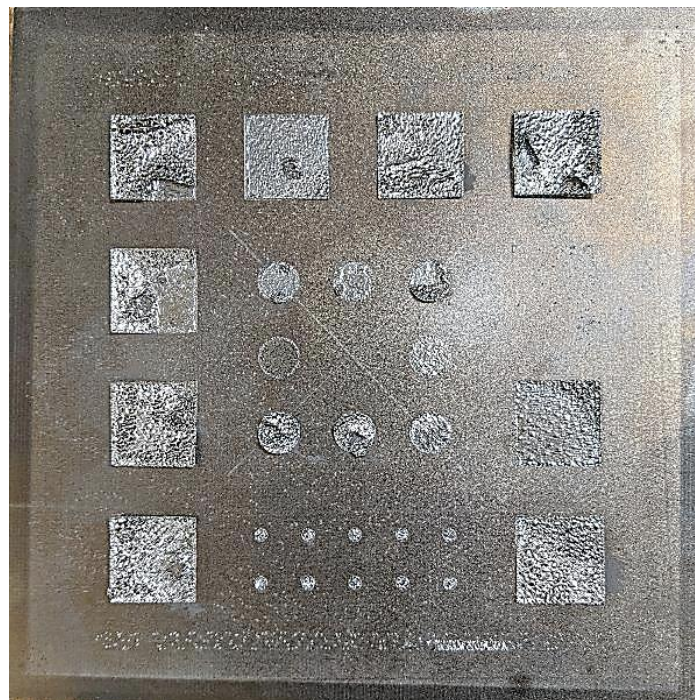


Build 2 Lattice Manufacturing Trial Build 5/1/21 (failure)

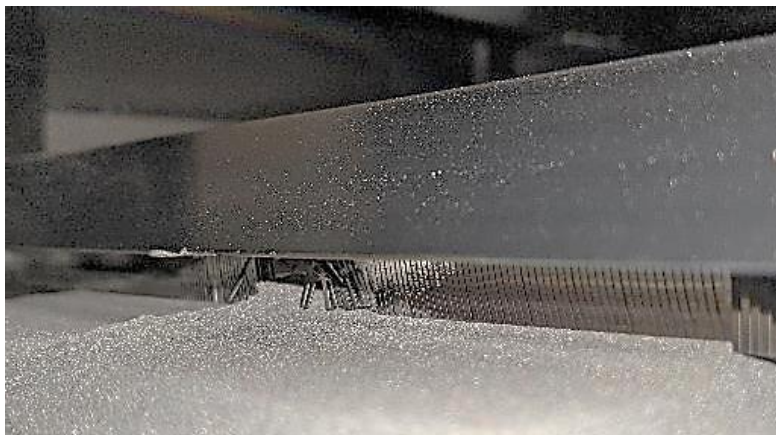




Build 3 Lattice Manufacturing Trial v1 Build 6/5/21 (failure)



Build 4 Lattice Manufacturing Trial v1 Build 6/12/21 (success, partial)



Build 5 Lattice Manufacturing Trial v2 Build 6/26/21 (success)

layers that “smoked” i.e. arc trips:

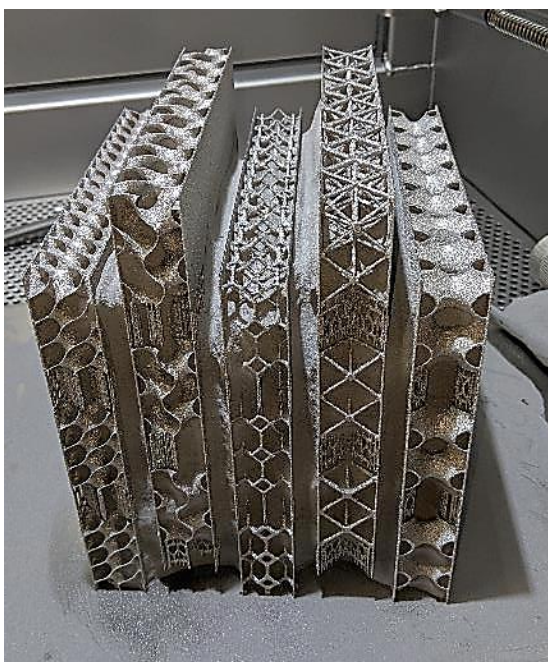
Time Stamp	Z-Level	Category	Message
6/27/2021 10:19:53 AM	104.3000 mm	RecoverableError	Arc Trip Warning
6/27/2021 9:40:03 AM	99.9500 mm	RecoverableError	Arc Trip Warning
6/27/2021 8:03:20 AM	89.9500 mm	RecoverableError	Arc Trip Warning
6/26/2021 5:09:17 PM	10.7000 mm	RecoverableError	Arc Trip Warning
6/26/2021 5:02:08 PM	9.9500 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:19:17 PM	5.3000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:15:10 PM	4.8000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:07:38 PM	3.8000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:03:46 PM	3.3000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:01:47 PM	3.0500 mm	RecoverableError	Arc Trip Warning

0 % Task 0 % Process stopped

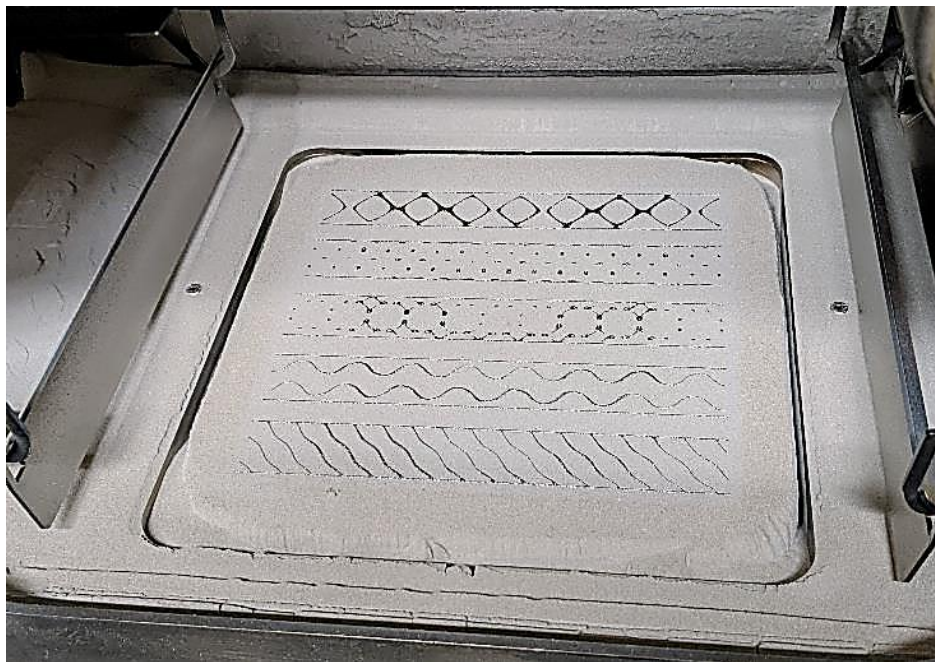
Time Stamp	Z-Level	Category	Message
6/26/2021 5:09:17 PM	10.7000 mm	RecoverableError	Arc Trip Warning
6/26/2021 5:02:08 PM	9.9500 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:19:17 PM	5.3000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:15:10 PM	4.8000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:07:38 PM	3.8000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:03:46 PM	3.3000 mm	RecoverableError	Arc Trip Warning
6/26/2021 4:01:47 PM	3.0500 mm	RecoverableError	Arc Trip Warning
6/26/2021 3:53:23 PM	1.9500 mm	RecoverableError	Arc Trip Warning
6/26/2021 3:11:59 PM	0.0500 mm	Information	Build Failed
6/26/2021 3:11:59 PM	0.0500 mm	Error	Grid too low error

0 % Task 0 % Process stopped

Build 6 Lattice Sandwich Beam Build 10/16/21 (success)



Build 7 Lattice Sandwich Beam Build 1/8/22 (partial success)



All Beam Specimen Lineup



Temperature Ramp Rate Summary Log

Preheat	Build #1 Line Config				Build #2 Lattice Config 1				Build #3 Lattice Config 1			
Beam Current	Start Temp	End Temp	Start Time	End Time	Start Temp	End Temp	Start Time	End Time	Start Temp	End Temp	Start Time	End Time
mA	°C	°C	h:mm:ss	h:mm:ss	°C	°C	h:mm:ss	h:mm:ss	°C	°C	h:mm:ss	h:mm:ss
1	26	32	0:00:00	0:07:51	-	-	-	-	30	40	0:00:00	0:16:28
10	31	70	0:10:22	0:15:26	26	70	0:00:00	0:05:54	40	100	0:19:07	0:29:45
20	79	135	0:17:05	0:20:49	76	101	0:06:13	0:08:56	102	161	0:30:03	0:33:38
25	151	186	0:22:12	0:25:00	121	191	0:09:17	0:12:33	164	241	0:33:58	0:40:27
30	219	336	0:26:49	0:30:50	204	256	0:12:46	0:15:48	246	362	0:40:40	0:44:41
35	352	486	0:32:12	0:52:32	292	404	0:16:06	0:19:20	360	511	0:44:54	1:04:44
40	484	601	0:53:26	0:56:49	416	554	0:19:39	0:37:51	510	620	1:05:03	1:15:24
45	591	650	0:57:34	0:59:47	578	609	0:38:10	0:39:37	624	671	1:15:38	1:18:30
48 (default)	623	730	1:01:01	1:39:21	634	730	0:39:55	0:43:02	674	730	1:18:40	1:32:16

Preheat	Build #4 Lattice config 1				Build #5 Lattice Config 2				Build #6 Lattice Config 3 beam			
Beam Current	Start Temp	End Temp	Start Time	End Time	Start Temp	End Temp	Start Time	End Time	Start Temp	End Temp	Start Time	End Time
mA	°C	°C	h:mm:ss	h:mm:ss	°C	°C	h:mm:ss	h:mm:ss	°C	°C	h:mm:ss	h:mm:ss
1	26	34	0:00:00	0:20:50	-	-	-	-	-	-	-	-
10	34	100	0:21:06	0:36:28	29	102	0:00:00	0:10:41	24	101	0:00:00	0:10:06
20	102	161	0:36:43	0:44:15	103	162	0:10:53	0:15:23	101	163	0:10:18	0:13:32
25	164	241	0:44:32	0:49:42	175	243	0:15:39	0:18:20	165	244	0:13:42	0:16:05
30	249	366	0:50:05	0:58:08	254	365	0:18:33	0:24:39	248	390	0:16:20	0:34:39
35	366	512	0:58:21	1:16:40	370	512	0:24:55	0:46:55	388	510	0:34:48	0:38:59
40	512	620	1:16:53	1:19:59	520	620	0:47:09	1:00:49	511	620	0:39:09	0:43:27
45	621	672	1:20:10	1:27:14	621	671	1:01:01	1:11:46	618	670	0:43:36	0:47:58
48 (default)	674	730	1:27:26	1:30:22	670	730	1:12:01	1:44:31	670	730	0:48:07	1:02:07

Preheat	Beam Current	mA	1	10	20	25	30	35	40	45	48 (default)
Build #1	Start Temp	°C	26	31	79	151	219	352	484	591	623
	End Temp	°C	32	70	135	186	336	486	601	650	730
	Start Time	h:mm:ss	0:00:00	0:10:22	0:17:05	0:22:12	0:26:49	0:32:12	0:53:26	0:57:34	1:01:01
	End Time	h:mm:ss	0:07:51	0:15:26	0:20:49	0:25:00	0:30:50	0:52:32	0:56:49	0:59:47	1:39:21
Build #2	Start Temp	°C	-	26	76	121	204	292	416	578	634
	End Temp	°C	-	70	101	191	256	404	554	609	730
	Start Time	h:mm:ss	-	0:00:00	0:06:13	0:09:17	0:12:46	0:16:06	0:19:39	0:38:10	0:39:55
	End Time	h:mm:ss	-	0:05:54	0:08:56	0:12:33	0:15:48	0:19:20	0:37:51	0:39:37	0:43:02
Build #3	Start Temp	°C	30	40	102	164	246	360	510	624	674
	End Temp	°C	40	100	161	241	362	511	620	671	730
	Start Time	h:mm:ss	0:00:00	0:19:07	0:30:03	0:33:58	0:40:40	0:44:54	1:05:03	1:15:38	1:18:40
	End Time	h:mm:ss	0:16:28	0:29:45	0:33:38	0:40:27	0:44:41	1:04:44	1:15:24	1:18:30	1:32:16
Build #4	Start Temp	°C	26	34	102	164	249	366	512	621	674
	End Temp	°C	34	100	161	241	366	512	620	672	730
	Start Time	h:mm:ss	0:00:00	0:21:06	0:36:43	0:44:32	0:50:05	0:58:21	1:16:53	1:20:10	1:27:26
	End Time	h:mm:ss	0:20:50	0:36:28	0:44:15	0:49:42	0:58:08	1:16:40	1:19:59	1:27:14	1:30:22

APPENDIX D

Tensile and compression coupon data and images from initial mechanical testing (Build 5)

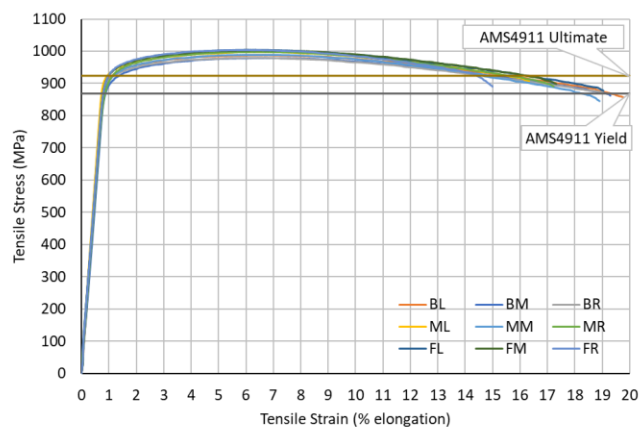
Specimen label	back-left		Specimen label	back-mid2	
Test	5		Test	2	
End date	7/27/2021		End date	7/27/2021	
Area	61.9321 mm ²		Area	61.9321 mm ²	
Length	25.4 mm		Length	25.4 mm	
Maximum Load	61178.55859 N		Maximum Load	60639.79297 N	
Tensile stress at Maximum Load	987.83282 MPa		Tensile stress at Maximum Load	979.13348 MPa	
Modulus (Automatic)	105872.6006 MPa		Modulus (Automatic)	100418.2512 MPa	
% Elongation	0.19883 mm/mm		% Elongation	0.19295 mm/mm	
Tensile stress at Break (Standard)	852.6015 MPa		Tensile stress at Break (Standard)	862.41028 MPa	
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.0665 mm/mm		Tensile strain (Extension) at Maximum Load	0.06675 mm/mm	

Specimen label	back-right		Specimen label	Middle Left	
Test	3		Test	4	
End date	7/27/2021		End date	7/27/2021	
Area	61.51435 mm ²		Area	61.9321 mm ²	
Length	25.4 mm		Length	25.4 mm	
Maximum Load	60273.19531 N		Maximum Load	61794.71484 N	
Tensile stress at Maximum Load	979.82336 MPa		Tensile stress at Maximum Load	997.78168 MPa	
Modulus (Automatic)	111531.8496 MPa		Modulus (Automatic)	109773.7327 MPa	
% Elongation	0.19019 mm/mm		% Elongation	0.16329 mm/mm	
Tensile stress at Break (Standard)	830.83777 MPa		Tensile stress at Break (Standard)	906.84784 MPa	
Tensile stress at Yield (Offset 0.2 %)	872.76676 MPa		Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.0645 mm/mm		Tensile strain (Extension) at Maximum Load	0.06741 mm/mm	

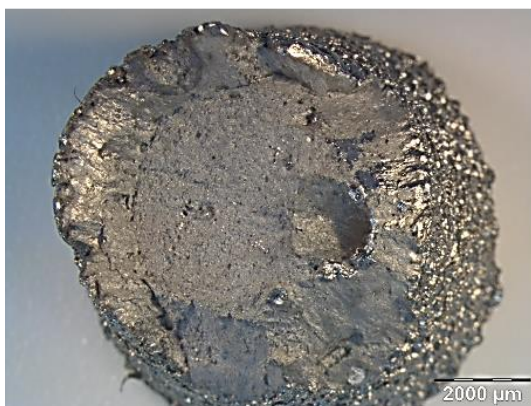
Specimen label	mid-mid		Specimen label	middle-right	
Test	8		Test	7	
End date	7/27/2021		End date	7/27/2021	
Area	62.21139 mm ²		Area	61.65344 mm ²	
Length	25.4 mm		Length	25.4 mm	
Maximum Load	61509.87109 N		Maximum Load	61506.02734 N	
Tensile stress at Maximum Load	988.72369 MPa		Tensile stress at Maximum Load	997.60895 MPa	
Modulus (Automatic)	98971.45801 MPa		Modulus (Automatic)	100822.7124 MPa	
% Elongation	0.18899 mm/mm		% Elongation	0.17223 mm/mm	
Tensile stress at Break (Standard)	845.3244 MPa		Tensile stress at Break (Standard)	890.83698 MPa	
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)	916.97653 MPa	
Tensile strain (Extension) at Maximum Load	0.06648 mm/mm		Tensile strain (Extension) at Maximum Load	0.06448 mm/mm	

Specimen label	front-left		Specimen label	front-mid	
Test	10		Test	6	
End date	7/27/2021		End date	7/27/2021	
Area	61.65344 mm ²		Area	61.9321 mm ²	
Length	25.4 mm		Length	25.4 mm	
Maximum Load	61923.74609 N		Maximum Load	62088.34375 N	
Tensile stress at Maximum Load	1004.38422 MPa		Tensile stress at Maximum Load	1002.52283 MPa	
Modulus (Automatic)	105718.3488 MPa		Modulus (Automatic)	104725.9455 MPa	
% Elongation	0.1904 mm/mm		% Elongation	0.17319 mm/mm	
Tensile stress at Break (Standard)	877.38934 MPa		Tensile stress at Break (Standard)	898.54822 MPa	
Tensile stress at Yield (Offset 0.2 %)	931.35982 MPa		Tensile stress at Yield (Offset 0.2 %)	932.50847 MPa	
Tensile strain (Extension) at Maximum Load	0.06057 mm/mm		Tensile strain (Extension) at Maximum Load	0.0691 mm/mm	

Specimen label	Front-Right	
Test	9	
End date	7/27/2021	
Area	61.9321	mm ²
Length	25.4	mm
Maximum Load	62158.86719	N
Tensile stress at Maximum Load	1003.66156	MPa
Modulus (Automatic)	104784.8185	MPa
% Elongation	0.14995	mm/mm
Tensile stress at Break (Standard)	890.36182	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.06229	mm/mm



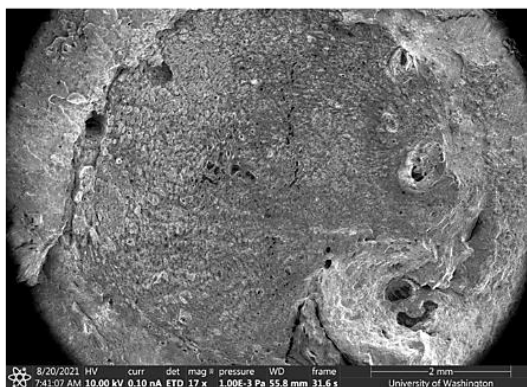
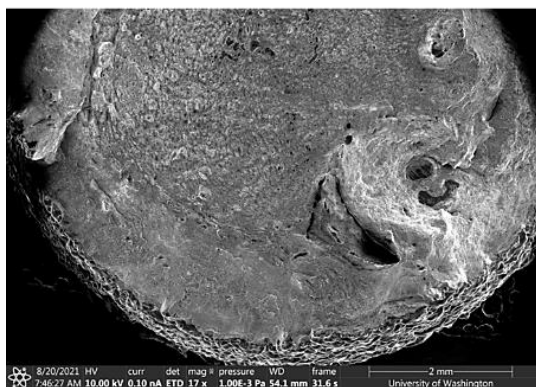
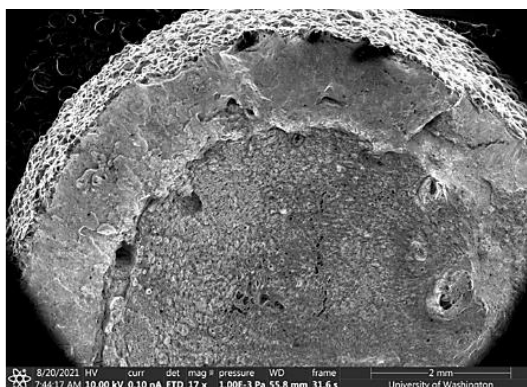
MM – Strain 18.9%



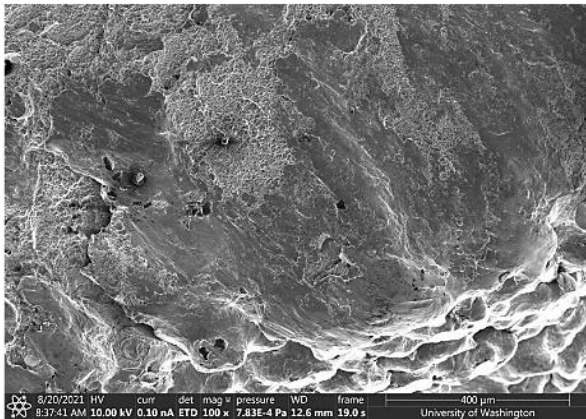
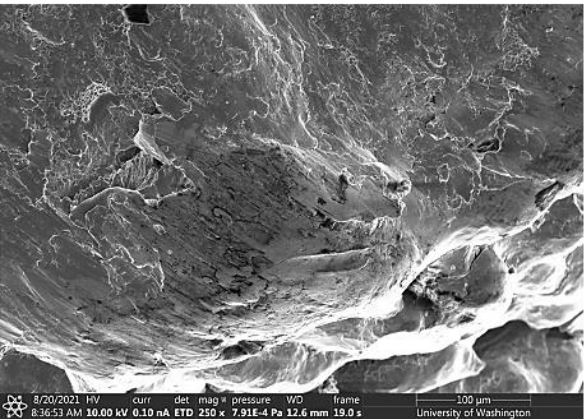
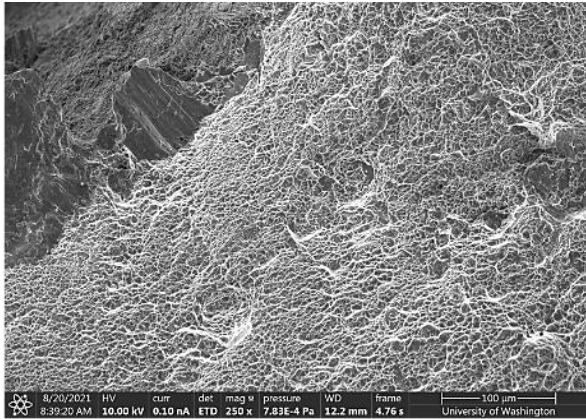
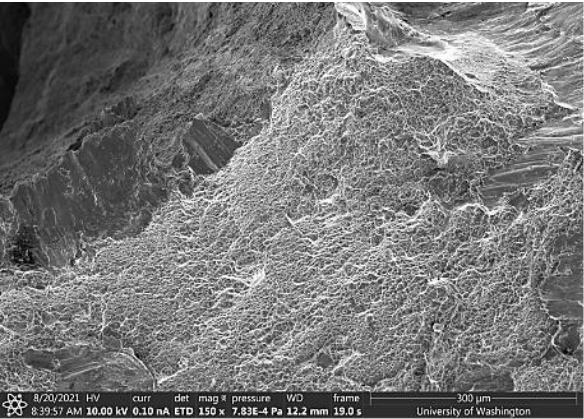
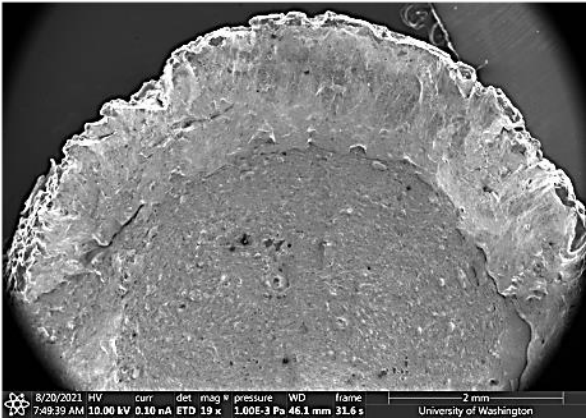
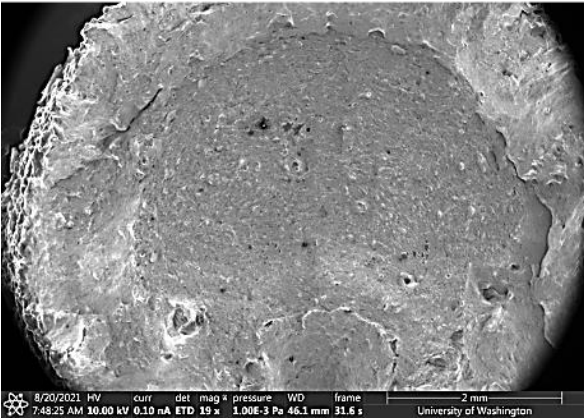
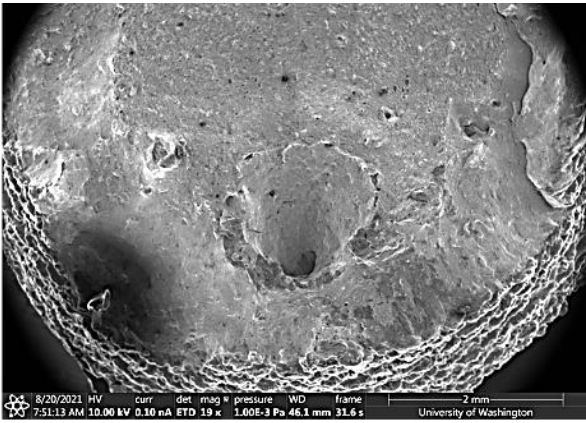
FR – Strain 15.5%

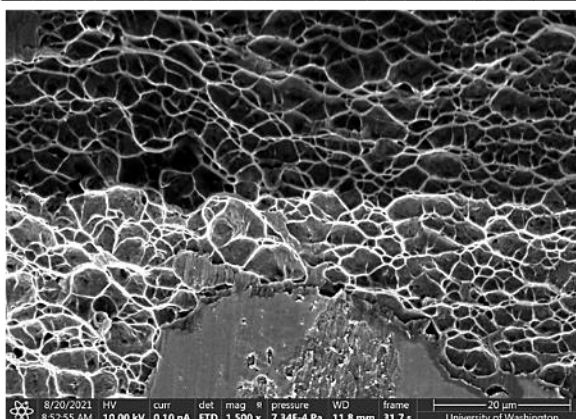
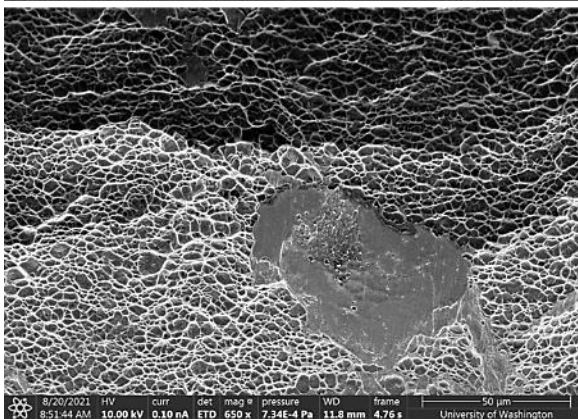
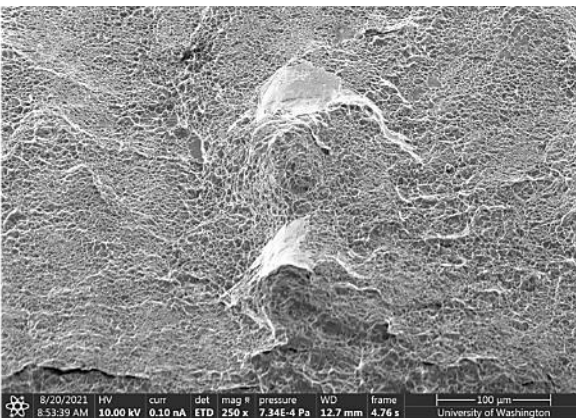
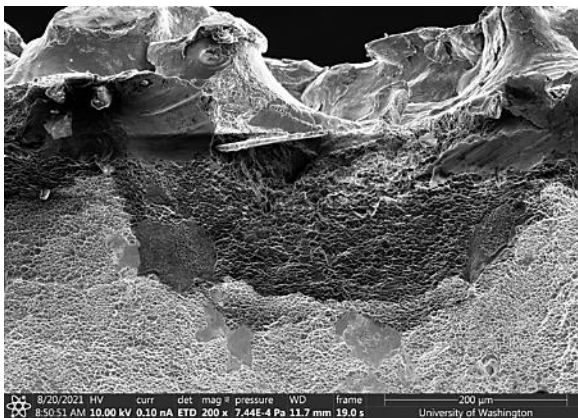
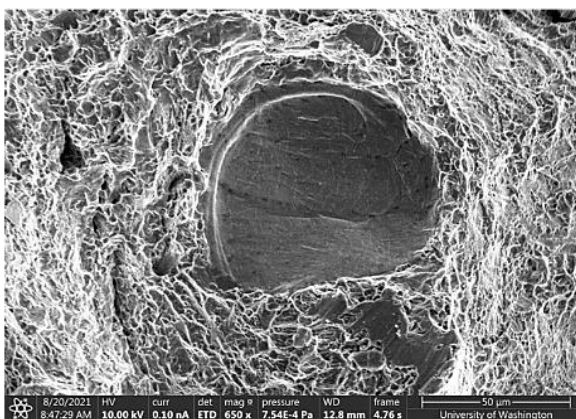
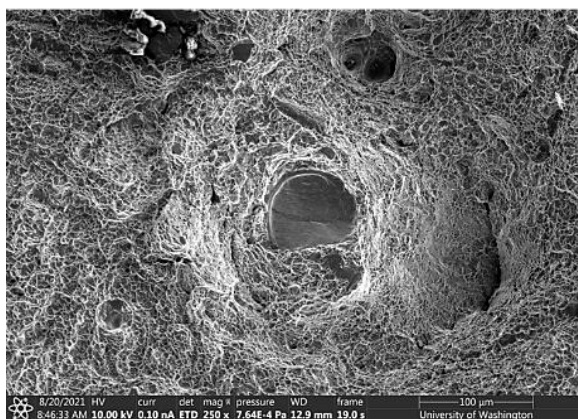


FR



MM





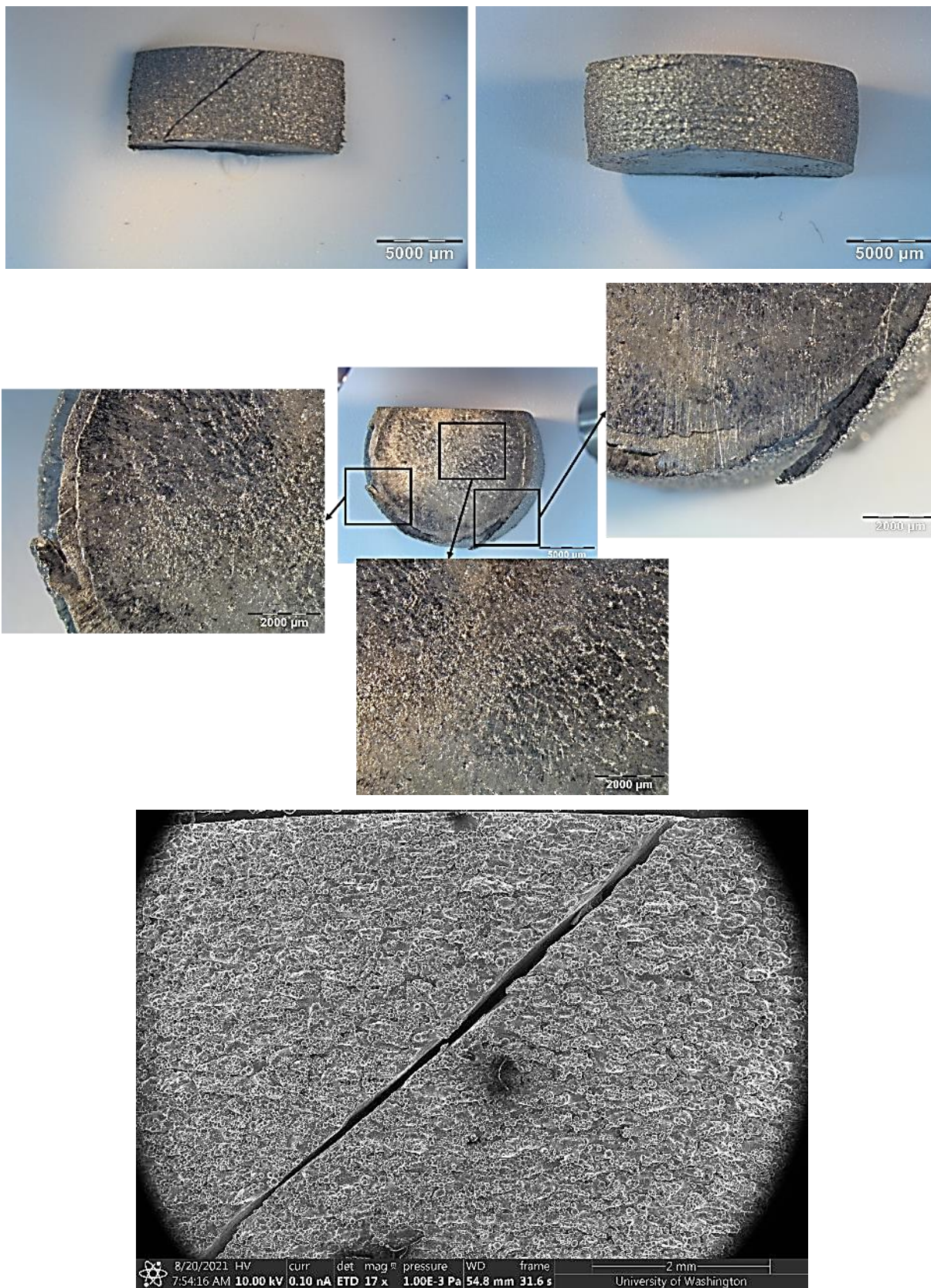
Specimen label	1	Specimen label	2
End date	7/27/2021	End date	7/27/2021
Diameter	12.55 mm	Diameter	12.55 mm
Area	123.70217 mm ²	Area	123.70217 mm ²
Anvil height	8.6 mm	Anvil height	8.6 mm
Maximum Load	243874.9844 N	Maximum Load	250494.4531 N
Compressive stress at Maximum Load	1971.46887 MPa	Compressive stress at Maximum Load	2024.98022 MPa
Modulus (Automatic Young's)	9281.36269 MPa	Modulus (Automatic Young's)	8253.91148 MPa
Compressive stress at Yield (Offset 0.2 %)	1037.12269 MPa	Compressive stress at Yield (Offset 0.2 %)	1054.72707 MPa
Rate 1	1.27 mm/min	Rate 1	1.27 mm/min

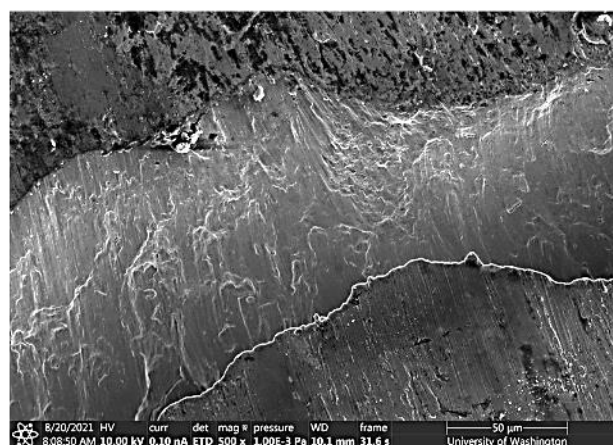
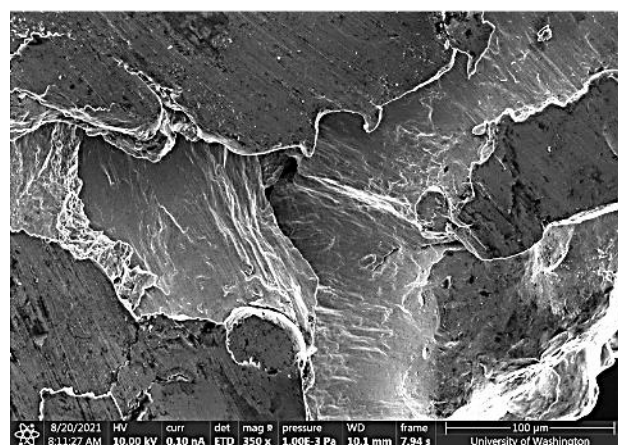
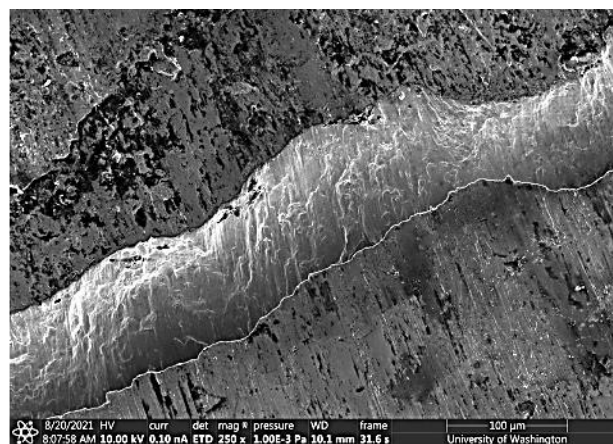
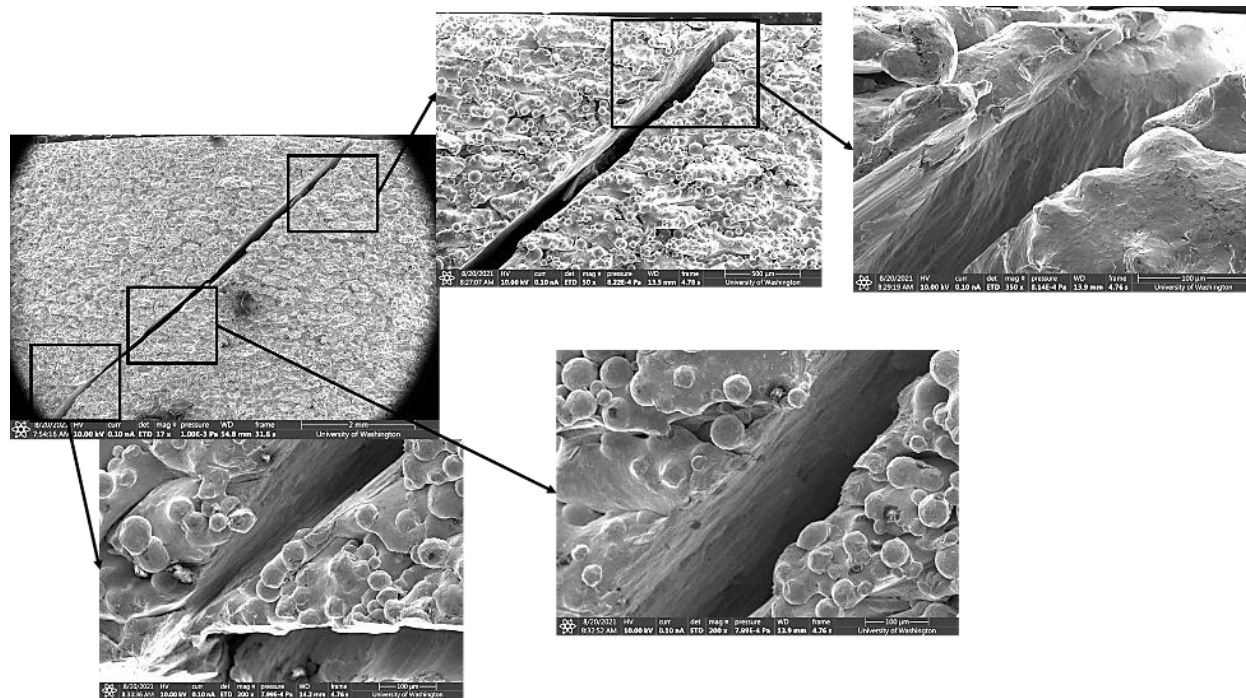
Specimen label	3	Specimen label	4
End date	7/27/2021	End date	7/27/2021
Diameter	12.55 mm	Diameter	12.55 mm
Area	123.70217 mm ²	Area	123.70217 mm ²
Anvil height	8.6 mm	Anvil height	8.6 mm
Maximum Load	250485.6563 N	Maximum Load	250476.3906 N
Compressive stress at Maximum Load	2024.90918 MPa	Compressive stress at Maximum Load	2024.83423 MPa
Modulus (Automatic Young's)	8792.96308 MPa	Modulus (Automatic Young's)	8387.81009 MPa
Compressive stress at Yield (Offset 0.2 %)	1079.59629 MPa	Compressive stress at Yield (Offset 0.2 %)	1015.4565 MPa
Rate 1	1.27 mm/min	Rate 1	1.27 mm/min

Specimen label	5	Specimen label	6
End date	7/27/2021	End date	7/27/2021
Diameter	12.55 mm	Diameter	12.55 mm
Area	123.70217 mm ²	Area	123.70217 mm ²
Anvil height	8.6 mm	Anvil height	8.6 mm
Maximum Load	250496.3594 N	Maximum Load	250498.9844 N
Compressive stress at Maximum Load	2024.99561 MPa	Compressive stress at Maximum Load	2025.01685 MPa
Modulus (Automatic Young's)	9185.99735 MPa	Modulus (Automatic Young's)	8576.32865 MPa
Compressive stress at Yield (Offset 0.2 %)	1068.32864 MPa	Compressive stress at Yield (Offset 0.2 %)	1091.1838 MPa
Rate 1	1.27 mm/min	Rate 1	1.27 mm/min

Specimen label	7	Specimen label	8
End date	7/27/2021	End date	7/27/2021
Diameter	12.55 mm	Diameter	12.55 mm
Area	123.70217 mm ²	Area	123.70217 mm ²
Anvil height	8.6 mm	Anvil height	8.6 mm
Maximum Load	250495.5156 N	Maximum Load	250480.7188 N
Compressive stress at Maximum Load	2024.98889 MPa	Compressive stress at Maximum Load	2024.86926 MPa
Modulus (Automatic Young's)	8702.07694 MPa	Modulus (Automatic Young's)	9032.16019 MPa
Compressive stress at Yield (Offset 0.2 %)	1069.30128 MPa	Compressive stress at Yield (Offset 0.2 %)	1071.27279 MPa
Rate 1	1.27 mm/min	Rate 1	1.27 mm/min

Specimen label	9	Specimen label	lattice
End date	7/27/2021	End date	7/27/2021
Diameter	12.55 mm	Diameter	8.1 mm
Area	123.70217 mm ²	Area	51.52997 mm ²
Anvil height	8.6 mm	Anvil height	8.6 mm
Maximum Load	250483.5156 N	Maximum Load	18180.51172 N
Compressive stress at Maximum Load	2024.89185 MPa	Compressive stress at Maximum Load	352.8143 MPa
Modulus (Automatic Young's)	8319.09681 MPa	Modulus (Automatic Young's)	5321.33279 MPa
Compressive stress at Yield (Offset 0.2 %)	1104.26239 MPa	Compressive stress at Yield (Offset 0.2 %)	288.51631 MPa
Rate 1	1.27 mm/min	Rate 1	1.27 mm/min







Tensile coupon data and images from final mechanical testing (Build 12)

Specimen label	L-G1		Specimen label	L-G2	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	70373.92188	N	Maximum Load	68920.0625	N
Tensile stress at Maximum Load	1106.20825	MPa	Tensile stress at Maximum Load	1083.35498	MPa
Modulus (Automatic)	110726.4425	MPa	Modulus (Automatic)	108668.4495	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.08567	mm/mm	% Elongation	0.08597	mm/mm
Tensile stress at Break (Standard)	1090.09131	MPa	Tensile stress at Break (Standard)	1073.00842	MPa
Tensile stress at Yield (Offset 0.2 %)	1032.11806	MPa	Tensile stress at Yield (Offset 0.2 %)	1011.09625	MPa
Tensile strain (Extension) at Maximum Load	0.05027	mm/mm	Tensile strain (Extension) at Maximum Load	0.05742	mm/mm

Specimen label	L-G3		Specimen label	L-G4	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	68528.83594	N	Maximum Load	68310.53125	N
Tensile stress at Maximum Load	1077.20532	MPa	Tensile stress at Maximum Load	1073.7738	MPa
Modulus (Automatic)	104956.8768	MPa	Modulus (Automatic)	108526.0992	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.11571	mm/mm	% Elongation	0.11709	mm/mm
Tensile stress at Break (Standard)	1048.26428	MPa	Tensile stress at Break (Standard)	1044.74207	MPa
Tensile stress at Yield (Offset 0.2 %)	1008.29876	MPa	Tensile stress at Yield (Offset 0.2 %)	1003.989	MPa
Tensile strain (Extension) at Maximum Load	0.06271	mm/mm	Tensile strain (Extension) at Maximum Load	0.05929	mm/mm

Specimen label	L-G5		Specimen label	L-G6	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	68574.76562	N	Maximum Load	69262.82031	N
Tensile stress at Maximum Load	1077.92725	MPa	Tensile stress at Maximum Load	1088.7428	MPa
Modulus (Automatic)	109564.9173	MPa	Modulus (Automatic)	110651.9364	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.11252	mm/mm	% Elongation	0.05016	mm/mm
Tensile stress at Break (Standard)	1053.65637	MPa	Tensile stress at Break (Standard)	1080.58459	MPa
Tensile stress at Yield (Offset 0.2 %)	1009.81028	MPa	Tensile stress at Yield (Offset 0.2 %)	1017.45589	MPa
Tensile strain (Extension) at Maximum Load	0.06236	mm/mm	Tensile strain (Extension) at Maximum Load	0.05242	mm/mm

Specimen label	L-G7		Specimen label	L-P1	
End date	12/2/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	70104.10938	N	Maximum Load	68386.23438	N
Tensile stress at Maximum Load	1101.96704	MPa	Tensile stress at Maximum Load	1074.96375	MPa
Modulus (Automatic)	113958.5921	MPa	Modulus (Automatic)	109769.6222	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.05371	mm/mm	% Elongation	0.02385	mm/mm
Tensile stress at Break (Standard)	1098.99011	MPa	Tensile stress at Break (Standard)	1072.01331	MPa
Tensile stress at Yield (Offset 0.2 %)	1030.99638	MPa	Tensile stress at Yield (Offset 0.2 %)	1020.27584	MPa
Tensile strain (Extension) at Maximum Load	0.0463	mm/mm	Tensile strain (Extension) at Maximum Load	0.02277	mm/mm

Specimen label	L-P2		Specimen label	L-P3	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	69810.13281	N	Maximum Load	70014.48438	N
Tensile stress at Maximum Load	1097.34595	MPa	Tensile stress at Maximum Load	1100.55823	MPa
Modulus (Automatic)	116086.9221	MPa	Modulus (Automatic)	112836.0154	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.03718	mm/mm	% Elongation	0.05027	mm/mm
Tensile stress at Break (Standard)	1074.00293	MPa	Tensile stress at Break (Standard)	1078.07043	MPa
Tensile stress at Yield (Offset 0.2 %)	1028.89937	MPa	Tensile stress at Yield (Offset 0.2 %)	1027.91203	MPa
Tensile strain (Extension) at Maximum Load	0.03313	mm/mm	Tensile strain (Extension) at Maximum Load	0.04672	mm/mm

Specimen label	L-P4		Specimen label	L-P5	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	69346.39844	N	Maximum Load	70098.03125	N
Tensile stress at Maximum Load	1090.05652	MPa	Tensile stress at Maximum Load	1101.87146	MPa
Modulus (Automatic)	111162.2393	MPa	Modulus (Automatic)	111090.4692	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.03424	mm/mm	% Elongation	0.05136	mm/mm
Tensile stress at Break (Standard)	1086.60242	MPa	Tensile stress at Break (Standard)	1096.87952	MPa
Tensile stress at Yield (Offset 0.2 %)	1024.18941	MPa	Tensile stress at Yield (Offset 0.2 %)	1031.18933	MPa
Tensile strain (Extension) at Maximum Load	0.03174	mm/mm	Tensile strain (Extension) at Maximum Load	0.04743	mm/mm

Specimen label	L-P6		Specimen label	L-P7	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	70416.30469	N	Maximum Load	70749.11719	N
Tensile stress at Maximum Load	1106.87439	MPa	Tensile stress at Maximum Load	1112.10596	MPa
Modulus (Automatic)	112830.013	MPa	Modulus (Automatic)	112133.9312	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.04411	mm/mm	% Elongation	0.02924	mm/mm
Tensile stress at Break (Standard)	1088.30298	MPa	Tensile stress at Break (Standard)	1109.44836	MPa
Tensile stress at Yield (Offset 0.2 %)	1035.68073	MPa	Tensile stress at Yield (Offset 0.2 %)	1051.8754	MPa
Tensile strain (Extension) at Maximum Load	0.0417	mm/mm	Tensile strain (Extension) at Maximum Load	0.02732	mm/mm

Specimen label	L-D1		Specimen label	L-D2	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	71402.60156	N	Maximum Load	69964.58594	N
Tensile stress at Maximum Load	1122.37805	MPa	Tensile stress at Maximum Load	1099.7738	MPa
Modulus (Automatic)	114064.8547	MPa	Modulus (Automatic)	111910.3381	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.03847	mm/mm	% Elongation	0.06381	mm/mm
Tensile stress at Break (Standard)	1121.34949	MPa	Tensile stress at Break (Standard)	1091.73193	MPa
Tensile stress at Yield (Offset 0.2 %)	1049.44726	MPa	Tensile stress at Yield (Offset 0.2 %)	1028.01944	MPa
Tensile strain (Extension) at Maximum Load	0.03701	mm/mm	Tensile strain (Extension) at Maximum Load	0.05173	mm/mm

Specimen label	L-D3		Specimen label	L-D4	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	69696.98438	N	Maximum Load	69464.19531	N
Tensile stress at Maximum Load	1095.56738	MPa	Tensile stress at Maximum Load	1091.9082	MPa
Modulus (Automatic)	109428.3153	MPa	Modulus (Automatic)	110052.2783	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.06855	mm/mm	% Elongation	0.05842	mm/mm
Tensile stress at Break (Standard)	1090.64612	MPa	Tensile stress at Break (Standard)	1089.63037	MPa
Tensile stress at Yield (Offset 0.2 %)	1025.19704	MPa	Tensile stress at Yield (Offset 0.2 %)	1023.57495	MPa
Tensile strain (Extension) at Maximum Load	0.05703	mm/mm	Tensile strain (Extension) at Maximum Load	0.05286	mm/mm

Specimen label	L-D5		Specimen label	L-D6	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	69661.52344	N	Maximum Load	69813.03906	N
Tensile stress at Maximum Load	1095.01001	MPa	Tensile stress at Maximum Load	1097.39172	MPa
Modulus (Automatic)	109126.1239	MPa	Modulus (Automatic)	110776.0269	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.06361	mm/mm	% Elongation	0.0399	mm/mm
Tensile stress at Break (Standard)	1092.32886	MPa	Tensile stress at Break (Standard)	1093.45068	MPa
Tensile stress at Yield (Offset 0.2 %)	1023.75292	MPa	Tensile stress at Yield (Offset 0.2 %)	1027.03623	MPa
Tensile strain (Extension) at Maximum Load	0.04924	mm/mm	Tensile strain (Extension) at Maximum Load	0.03859	mm/mm

Specimen label	L-D7		Specimen label	L-O1	
End date	12/2/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	70831.07031	N	Maximum Load	70006.13281	N
Tensile stress at Maximum Load	1113.39417	MPa	Tensile stress at Maximum Load	1100.42688	MPa
Modulus (Automatic)	112185.7078	MPa	Modulus (Automatic)	105176.6237	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.04422	mm/mm	% Elongation	0.03169	mm/mm
Tensile stress at Break (Standard)	1109.55176	MPa	Tensile stress at Break (Standard)	1099.43787	MPa
Tensile stress at Yield (Offset 0.2 %)	1042.62521	MPa	Tensile stress at Yield (Offset 0.2 %)	1035.32133	MPa
Tensile strain (Extension) at Maximum Load	0.03967	mm/mm	Tensile strain (Extension) at Maximum Load	0.03065	mm/mm

Specimen label	L-O2		Specimen label	L-O3	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	69979.13281	N	Maximum Load	69244.82031	N
Tensile stress at Maximum Load	1100.00256	MPa	Tensile stress at Maximum Load	1088.45984	MPa
Modulus (Automatic)	116562.5071	MPa	Modulus (Automatic)	130017.4304	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.07824	mm/mm	% Elongation	0.0292	mm/mm
Tensile stress at Break (Standard)	1070.68445	MPa	Tensile stress at Break (Standard)	1063.88806	MPa
Tensile stress at Yield (Offset 0.2 %)	1043.20573	MPa	Tensile stress at Yield (Offset 0.2 %)	990.31501	MPa
Tensile strain (Extension) at Maximum Load	0.05264	mm/mm	Tensile strain (Extension) at Maximum Load	0.00497	mm/mm

Specimen label	L-O4		Specimen label	L-O5	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	69220.92969	N	Maximum Load	69409.58594	N
Tensile stress at Maximum Load	1088.08435	MPa	Tensile stress at Maximum Load	1091.0498	MPa
Modulus (Automatic)	116337.416	MPa	Modulus (Automatic)	110069.3881	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.06088	mm/mm	% Elongation	0.09288	mm/mm
Tensile stress at Break (Standard)	1073.04651	MPa	Tensile stress at Break (Standard)	1080.04993	MPa
Tensile stress at Yield (Offset 0.2 %)	1016.4846	MPa	Tensile stress at Yield (Offset 0.2 %)	1022.26543	MPa
Tensile strain (Extension) at Maximum Load	0.04984	mm/mm	Tensile strain (Extension) at Maximum Load	0.06275	mm/mm

Specimen label	L-O6		Specimen label	L-O7	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	70122.23438	N	Maximum Load	70699.59375	N
Tensile stress at Maximum Load	1102.25195	MPa	Tensile stress at Maximum Load	1111.32739	MPa
Modulus (Automatic)	120507.0902	MPa	Modulus (Automatic)	112993.4526	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.06733	mm/mm	% Elongation	0.03212	mm/mm
Tensile stress at Break (Standard)	1097.46997	MPa	Tensile stress at Break (Standard)	1108.83618	MPa
Tensile stress at Yield (Offset 0.2 %)	1032.03544	MPa	Tensile stress at Yield (Offset 0.2 %)	1045.50165	MPa
Tensile strain (Extension) at Maximum Load	0.04984	mm/mm	Tensile strain (Extension) at Maximum Load	0.03022	mm/mm

Specimen label	L-K1		Specimen label	L-K2	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	70984.16406	N	Maximum Load	69464.45312	N
Tensile stress at Maximum Load	1115.80066	MPa	Tensile stress at Maximum Load	1091.91223	MPa
Modulus (Automatic)	111966.2006	MPa	Modulus (Automatic)	109684.692	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.05516	mm/mm	% Elongation	0.05995	mm/mm
Tensile stress at Break (Standard)	1112.91101	MPa	Tensile stress at Break (Standard)	1086.60205	MPa
Tensile stress at Yield (Offset 0.2 %)	1041.80435	MPa	Tensile stress at Yield (Offset 0.2 %)	1022.08311	MPa
Tensile strain (Extension) at Maximum Load	0.04513	mm/mm	Tensile strain (Extension) at Maximum Load	0.04903	mm/mm

Specimen label	L-K3		Specimen label	L-K4	
End date	12/2/2022		End date	12/2/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	68814.01562	N	Maximum Load	68288.375	N
Tensile stress at Maximum Load	1081.68799	MPa	Tensile stress at Maximum Load	1073.42542	MPa
Modulus (Automatic)	109147.9312	MPa	Modulus (Automatic)	110394.7036	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.10518	mm/mm	% Elongation	0.10604	mm/mm
Tensile stress at Break (Standard)	1051.63171	MPa	Tensile stress at Break (Standard)	1055.90808	MPa
Tensile stress at Yield (Offset 0.2 %)	1012.62438	MPa	Tensile stress at Yield (Offset 0.2 %)	1000.16541	MPa
Tensile strain (Extension) at Maximum Load	0.06349	mm/mm	Tensile strain (Extension) at Maximum Load	0.06305	mm/mm

Specimen label	L-K5		Specimen label	L-K6	
End date	12/2/2022		End date	12/3/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	68649.96875	N	Maximum Load	69414.86719	N
Tensile stress at Maximum Load	1079.10938	MPa	Tensile stress at Maximum Load	1091.13281	MPa
Modulus (Automatic)	106520.9369	MPa	Modulus (Automatic)	111160.6898	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.10298	mm/mm	% Elongation	0.05172	mm/mm
Tensile stress at Break (Standard)	1062.57056	MPa	Tensile stress at Break (Standard)	1072.12341	MPa
Tensile stress at Yield (Offset 0.2 %)	1007.21522	MPa	Tensile stress at Yield (Offset 0.2 %)	1022.08568	MPa
Tensile strain (Extension) at Maximum Load	0.06226	mm/mm	Tensile strain (Extension) at Maximum Load	0.04917	mm/mm

Specimen label	L-K7		Specimen label	A-G1	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	70550.53906	N	Maximum Load	64342.32031	N
Tensile stress at Maximum Load	1108.9845	MPa	Tensile stress at Maximum Load	1011.3974	MPa
Modulus (Automatic)	107407.911	MPa	Modulus (Automatic)	106631.8078	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.05167	mm/mm	% Elongation	0.17897	mm/mm
Tensile stress at Break (Standard)	1107.34229	MPa	Tensile stress at Break (Standard)	870.44421	MPa
Tensile stress at Yield (Offset 0.2 %)	1035.34032	MPa	Tensile stress at Yield (Offset 0.2 %)	940.43924	MPa
Tensile strain (Extension) at Maximum Load	0.04671	mm/mm	Tensile strain (Extension) at Maximum Load	0.05902	mm/mm

Specimen label	A-G2		Specimen label	A-G3	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	62904.6875	N	Maximum Load	62504.64844	N
Tensile stress at Maximum Load	988.79926	MPa	Tensile stress at Maximum Load	982.51105	MPa
Modulus (Automatic)	103873.036	MPa	Modulus (Automatic)	101234.5148	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.18137	mm/mm	% Elongation	0.1793	mm/mm
Tensile stress at Break (Standard)	868.68951	MPa	Tensile stress at Break (Standard)	836.68066	MPa
Tensile stress at Yield (Offset 0.2 %)	915.43204	MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.06345	mm/mm	Tensile strain (Extension) at Maximum Load	0.06874	mm/mm

Specimen label	A-G4		Specimen label	A-G5	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	62149.34766	N	Maximum Load	62397.58594	N
Tensile stress at Maximum Load	976.92603	MPa	Tensile stress at Maximum Load	980.82812	MPa
Modulus (Automatic)	103062.4325	MPa	Modulus (Automatic)	105347.24	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.18637	mm/mm	% Elongation	0.17912	mm/mm
Tensile stress at Break (Standard)	857.78864	MPa	Tensile stress at Break (Standard)	863.65344	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.06805	mm/mm	Tensile strain (Extension) at Maximum Load	0.06886	mm/mm

Specimen label	A-G6		Specimen label	A-G7	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	63266.82812	N	Maximum Load	64004.79297	N
Tensile stress at Maximum Load	994.49176	MPa	Tensile stress at Maximum Load	1006.0918	MPa
Modulus (Automatic)	109440.5238	MPa	Modulus (Automatic)	106283.8668	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.14674	mm/mm	% Elongation	0.14702	mm/mm
Tensile stress at Break (Standard)	866.12891	MPa	Tensile stress at Break (Standard)	917.30737	MPa
Tensile stress at Yield (Offset 0.2 %)	929.37007	MPa	Tensile stress at Yield (Offset 0.2 %)	935.4387	MPa
Tensile strain (Extension) at Maximum Load	0.06105	mm/mm	Tensile strain (Extension) at Maximum Load	0.061	mm/mm

Specimen label	A-P1		Specimen label	A-P2	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	64105.69531	N	Maximum Load	64407.75781	N
Tensile stress at Maximum Load	1007.67792	MPa	Tensile stress at Maximum Load	1012.42603	MPa
Modulus (Automatic)	106186.9886	MPa	Modulus (Automatic)	107154.352	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.0767	mm/mm	% Elongation	0.09921	mm/mm
Tensile stress at Break (Standard)	978.90729	MPa	Tensile stress at Break (Standard)	966.71118	MPa
Tensile stress at Yield (Offset 0.2 %)	934.89339	MPa	Tensile stress at Yield (Offset 0.2 %)	942.9587	MPa
Tensile strain (Extension) at Maximum Load	0.05406	mm/mm	Tensile strain (Extension) at Maximum Load	0.06114	mm/mm

Specimen label	A-P3		Specimen label	A-P4	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	63747.89062	N	Maximum Load	63643.71484	N
Tensile stress at Maximum Load	1002.05359	MPa	Tensile stress at Maximum Load	1000.41602	MPa
Modulus (Automatic)	108299.0087	MPa	Modulus (Automatic)	105260.4565	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.10717	mm/mm	% Elongation	0.13569	mm/mm
Tensile stress at Break (Standard)	906.39062	MPa	Tensile stress at Break (Standard)	926.22681	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)	353.60665	MPa
Tensile strain (Extension) at Maximum Load	0.06287	mm/mm	Tensile strain (Extension) at Maximum Load	0.06463	mm/mm

Specimen label	A-P5		Specimen label	A-P6	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	63792.96484	N	Maximum Load	64071.83594	N
Tensile stress at Maximum Load	1002.76208	MPa	Tensile stress at Maximum Load	1007.14569	MPa
Modulus (Automatic)	107091.4481	MPa	Modulus (Automatic)	112980.8102	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.13329	mm/mm	% Elongation	0.15229	mm/mm
Tensile stress at Break (Standard)	911.16376	MPa	Tensile stress at Break (Standard)	910.52625	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.0679	mm/mm	Tensile strain (Extension) at Maximum Load	0.06123	mm/mm

Specimen label	A-P7		Specimen label	A-D1	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	64823.84375	N	Maximum Load	64943.39453	N
Tensile stress at Maximum Load	1018.96649	MPa	Tensile stress at Maximum Load	1020.8457	MPa
Modulus (Automatic)	110713.2611	MPa	Modulus (Automatic)	110052.1471	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.10894	mm/mm	% Elongation	0.08789	mm/mm
Tensile stress at Break (Standard)	962.57666	MPa	Tensile stress at Break (Standard)	969.89166	MPa
Tensile stress at Yield (Offset 0.2 %)	950.94065	MPa	Tensile stress at Yield (Offset 0.2 %)	947.20161	MPa
Tensile strain (Extension) at Maximum Load	0.0565	mm/mm	Tensile strain (Extension) at Maximum Load	0.05173	mm/mm

Specimen label	A-D2		Specimen label	A-D3	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	64065.0625	N	Maximum Load	63914.25391	N
Tensile stress at Maximum Load	1007.03918	MPa	Tensile stress at Maximum Load	1004.66864	MPa
Modulus (Automatic)	107357.6326	MPa	Modulus (Automatic)	105417.4328	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.09288	mm/mm	% Elongation	0.13633	mm/mm
Tensile stress at Break (Standard)	960.85089	MPa	Tensile stress at Break (Standard)	932.81891	MPa
Tensile stress at Yield (Offset 0.2 %)	937.11228	MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.0612	mm/mm	Tensile strain (Extension) at Maximum Load	0.0643	mm/mm

Specimen label	A-D4		Specimen label	A-D5	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	63693.04688	N	Maximum Load	63649.48438	N
Tensile stress at Maximum Load	1001.19147	MPa	Tensile stress at Maximum Load	1000.50671	MPa
Modulus (Automatic)	106321.9934	MPa	Modulus (Automatic)	104589.2523	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.07324	mm/mm	% Elongation	0.10922	mm/mm
Tensile stress at Break (Standard)	922.9942	MPa	Tensile stress at Break (Standard)	947.23712	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)	930.20437	MPa
Tensile strain (Extension) at Maximum Load	0.06306	mm/mm	Tensile strain (Extension) at Maximum Load	0.06182	mm/mm

Specimen label	A-D6		Specimen label	A-D7	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	64088.16016	N	Maximum Load	64639.96484	N
Tensile stress at Maximum Load	1007.40228	MPa	Tensile stress at Maximum Load	1016.07611	MPa
Modulus (Automatic)	104436.387	MPa	Modulus (Automatic)	109917.9141	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.09467	mm/mm	% Elongation	0.10329	mm/mm
Tensile stress at Break (Standard)	967.55219	MPa	Tensile stress at Break (Standard)	972.08582	MPa
Tensile stress at Yield (Offset 0.2 %)	936.52048	MPa	Tensile stress at Yield (Offset 0.2 %)	946.32823	MPa
Tensile strain (Extension) at Maximum Load	0.06062	mm/mm	Tensile strain (Extension) at Maximum Load	0.05755	mm/mm

Specimen label	A-O1		Specimen label	A-O2	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	64141.57031	N	Maximum Load	63793.12109	N
Tensile stress at Maximum Load	1008.24182	MPa	Tensile stress at Maximum Load	1002.76453	MPa
Modulus (Automatic)	107495.0851	MPa	Modulus (Automatic)	105968.1631	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.09718	mm/mm	% Elongation	0.17669	mm/mm
Tensile stress at Break (Standard)	956.47382	MPa	Tensile stress at Break (Standard)	886.67554	MPa
Tensile stress at Yield (Offset 0.2 %)	936.31102	MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.05487	mm/mm	Tensile strain (Extension) at Maximum Load	0.0676	mm/mm

Specimen label	A-O3		Specimen label	A-O4	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	62787.68359	N	Maximum Load	62756.41016	N
Tensile stress at Maximum Load	986.96008	MPa	Tensile stress at Maximum Load	986.46844	MPa
Modulus (Automatic)	102234.5141	MPa	Modulus (Automatic)	100905.1697	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.03119	mm/mm	% Elongation	0.15114	mm/mm
Tensile stress at Break (Standard)	883.04797	MPa	Tensile stress at Break (Standard)	828.56232	MPa
Tensile stress at Yield (Offset 0.2 %)	946.96876	MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.06693	mm/mm	Tensile strain (Extension) at Maximum Load	0.07063	mm/mm

Specimen label	A-O5		Specimen label	A-O6	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	62802.48438	N	Maximum Load	63504.93359	N
Tensile stress at Maximum Load	987.19269	MPa	Tensile stress at Maximum Load	998.2345	MPa
Modulus (Automatic)	106739.0387	MPa	Modulus (Automatic)	105647.7326	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.19592	mm/mm	% Elongation	0.15191	mm/mm
Tensile stress at Break (Standard)	836.18005	MPa	Tensile stress at Break (Standard)	900.07056	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.06447	mm/mm	Tensile strain (Extension) at Maximum Load	0.06602	mm/mm

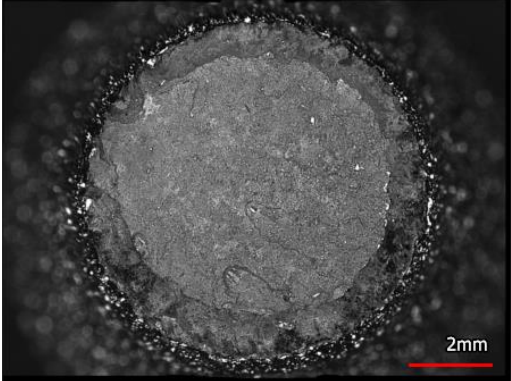
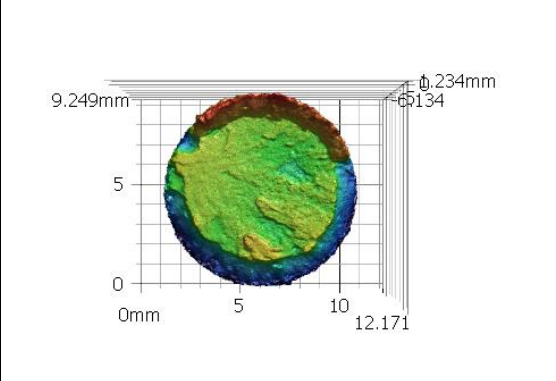
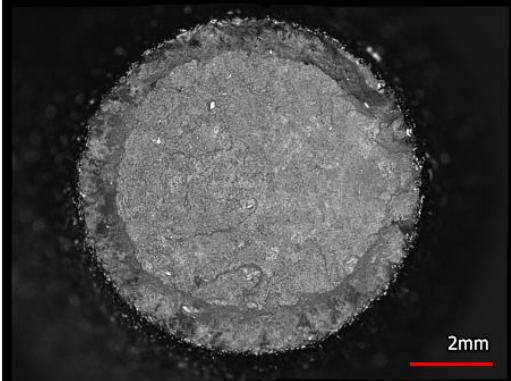
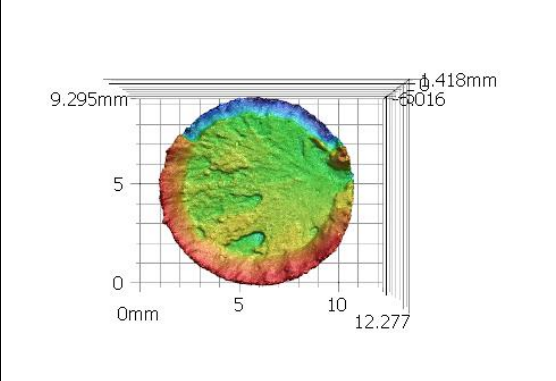
Specimen label	A-O7		Specimen label	A-K1	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	64310.72266	N	Maximum Load	64409.32422	N
Tensile stress at Maximum Load	1010.9007	MPa	Tensile stress at Maximum Load	1012.45062	MPa
Modulus (Automatic)	110063.6497	MPa	Modulus (Automatic)	103320.7766	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.12313	mm/mm	% Elongation	0.13767	mm/mm
Tensile stress at Break (Standard)	932.22797	MPa	Tensile stress at Break (Standard)	913.05939	MPa
Tensile stress at Yield (Offset 0.2 %)	943.46858	MPa	Tensile stress at Yield (Offset 0.2 %)	941.13778	MPa
Tensile strain (Extension) at Maximum Load	0.05788	mm/mm	Tensile strain (Extension) at Maximum Load	0.06337	mm/mm

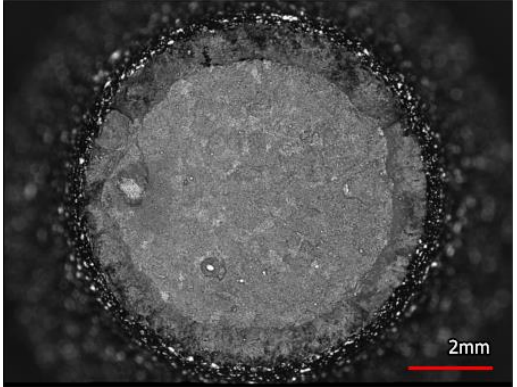
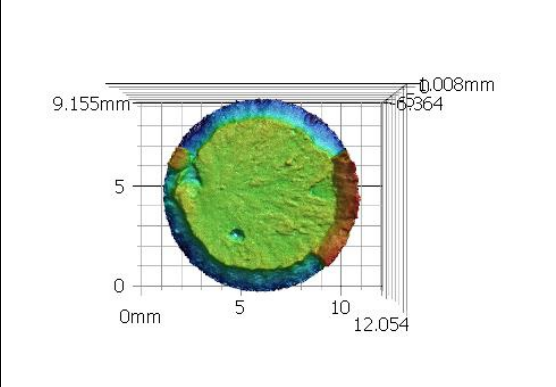
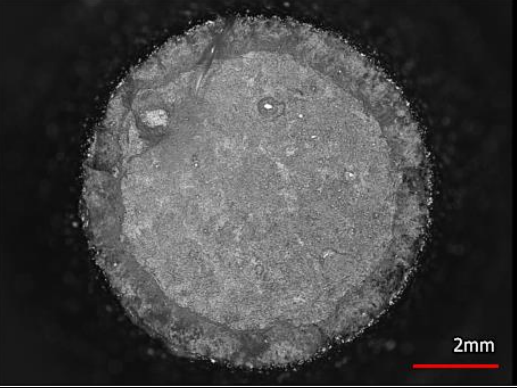
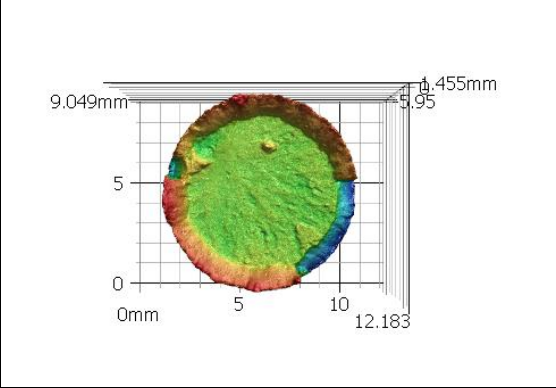
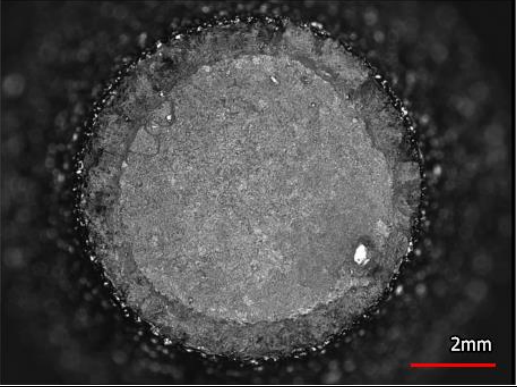
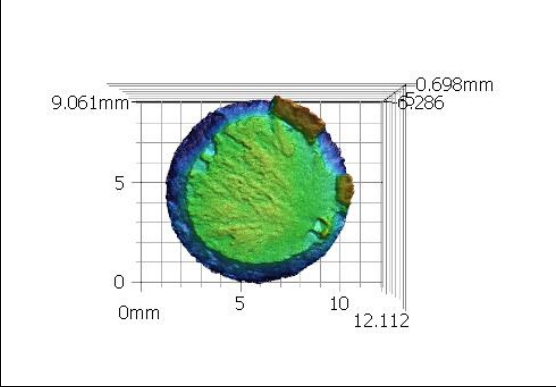
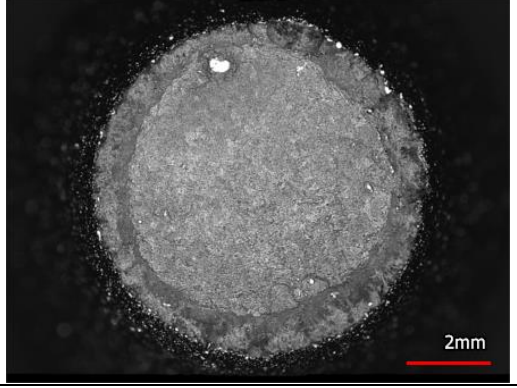
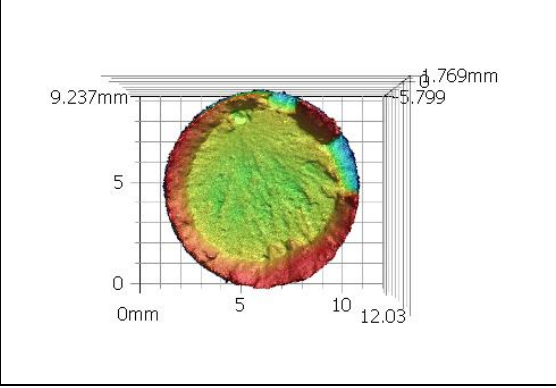
Specimen label	A-K2		Specimen label	A-K3	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	63095.42188	N	Maximum Load	62467.87109	N
Tensile stress at Maximum Load	991.79742	MPa	Tensile stress at Maximum Load	981.93292	MPa
Modulus (Automatic)	104848.1418	MPa	Modulus (Automatic)	104475.4402	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.08813	mm/mm	% Elongation	0.15061	mm/mm
Tensile stress at Break (Standard)	894.75623	MPa	Tensile stress at Break (Standard)	870.43219	MPa
Tensile stress at Yield (Offset 0.2 %)	921.09208	MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.06238	mm/mm	Tensile strain (Extension) at Maximum Load	0.06665	mm/mm

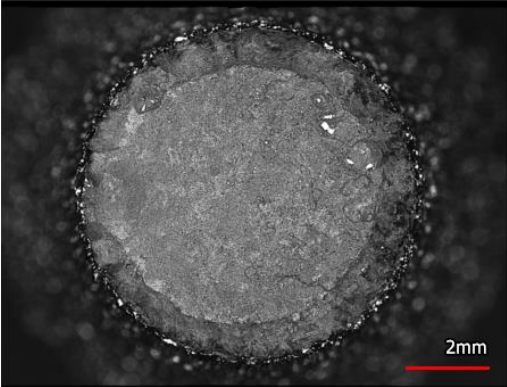
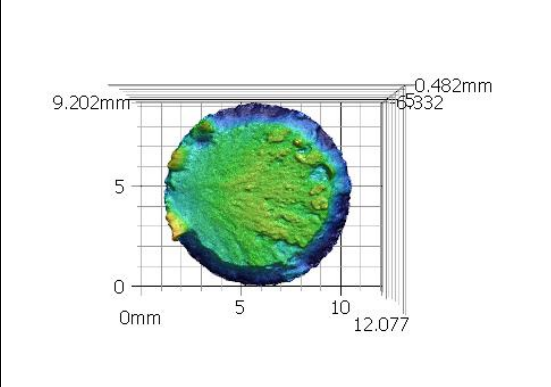
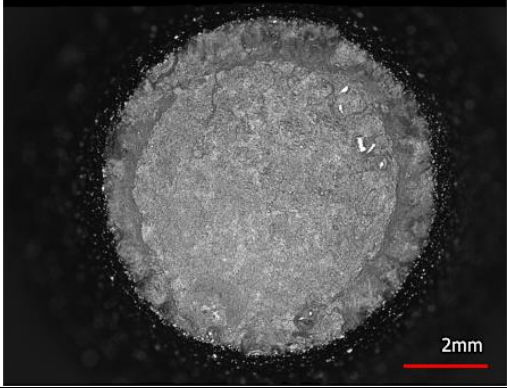
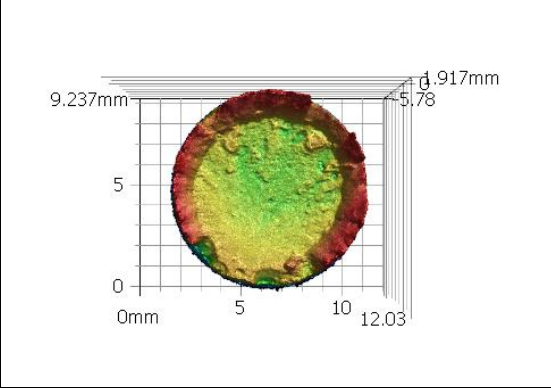
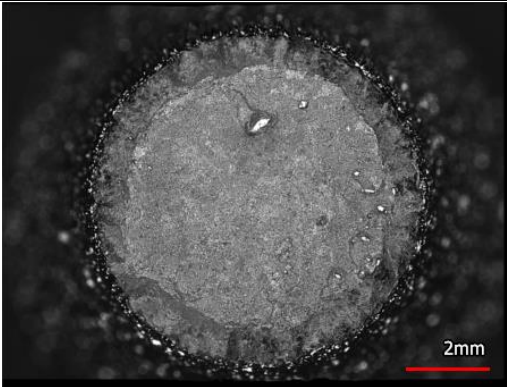
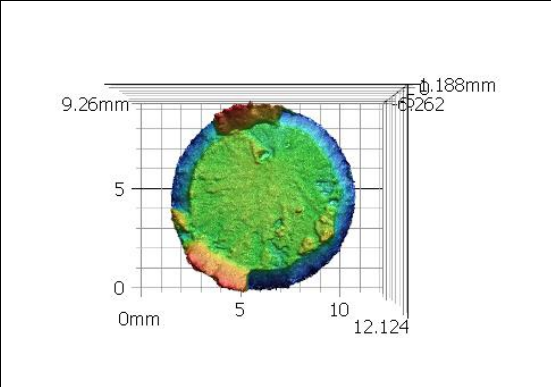
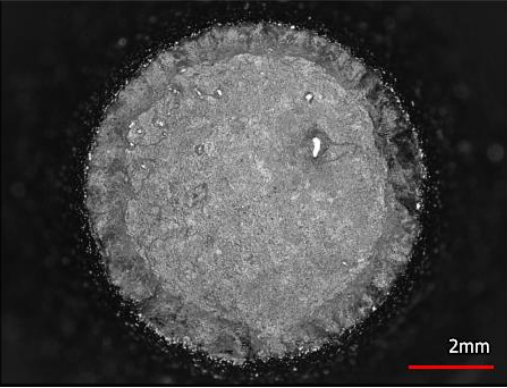
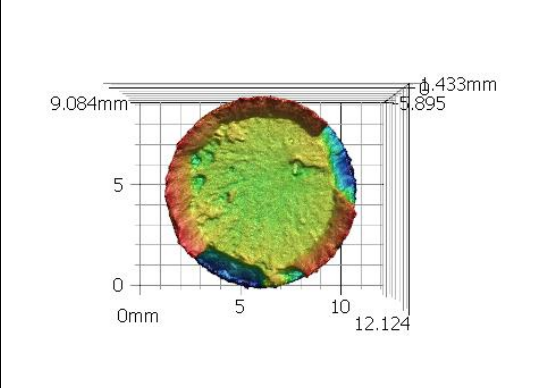
Specimen label	A-K4		Specimen label	A-K5	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm^2	Area	63.61725	mm^2
Length	25.4	mm	Length	25.4	mm
Maximum Load	62054.78516	N	Maximum Load	62322.46094	N
Tensile stress at Maximum Load	975.43964	MPa	Tensile stress at Maximum Load	979.64722	MPa
Modulus (Automatic)	104100.7655	MPa	Modulus (Automatic)	102580.2815	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.1739	mm/mm	% Elongation	0.15023	mm/mm
Tensile stress at Break (Standard)	871.17719	MPa	Tensile stress at Break (Standard)	847.27637	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)		MPa
Tensile strain (Extension) at Maximum Load	0.0695	mm/mm	Tensile strain (Extension) at Maximum Load	0.06847	mm/mm

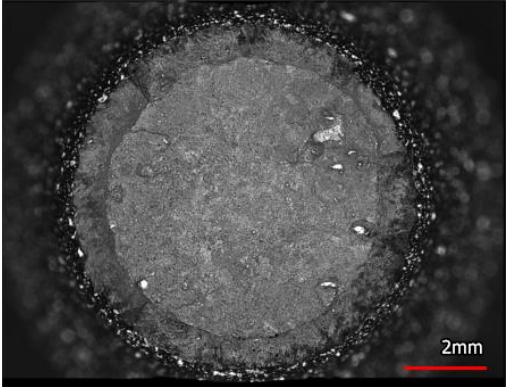
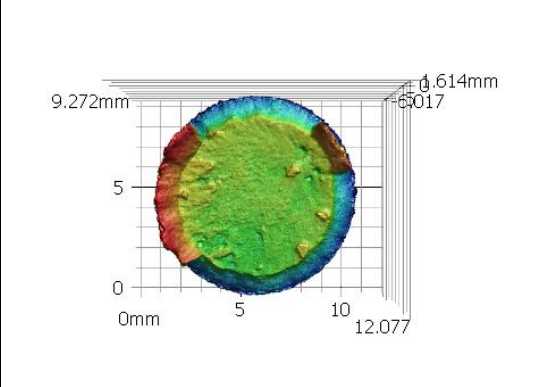
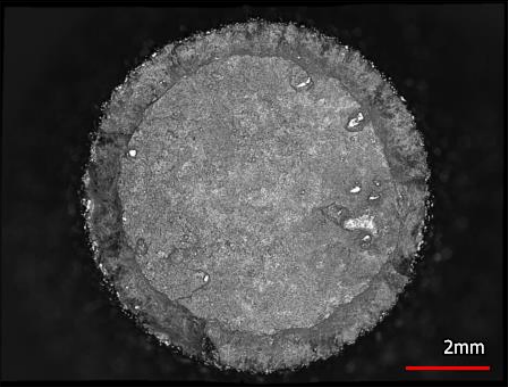
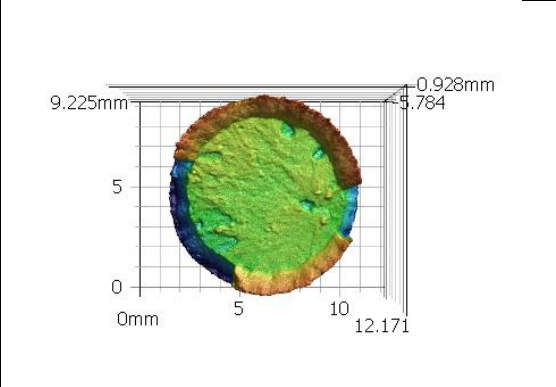
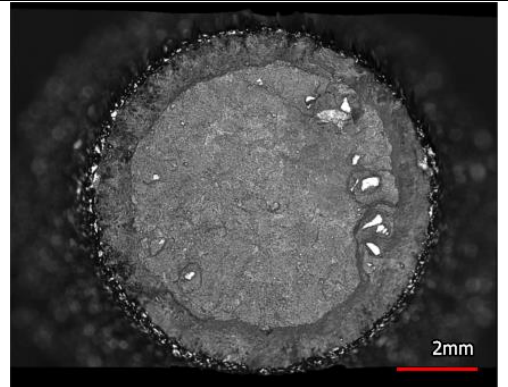
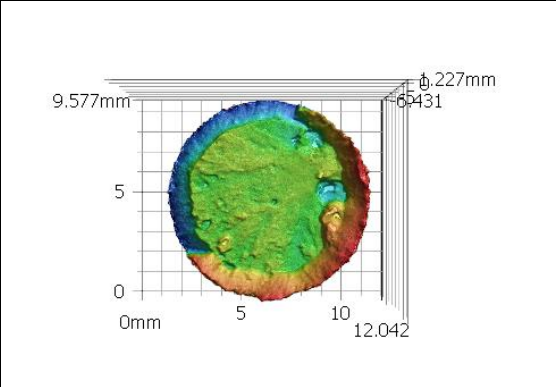
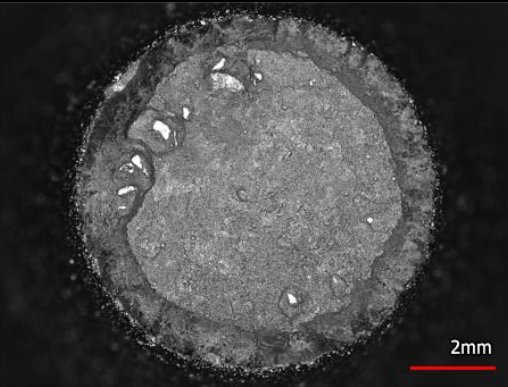
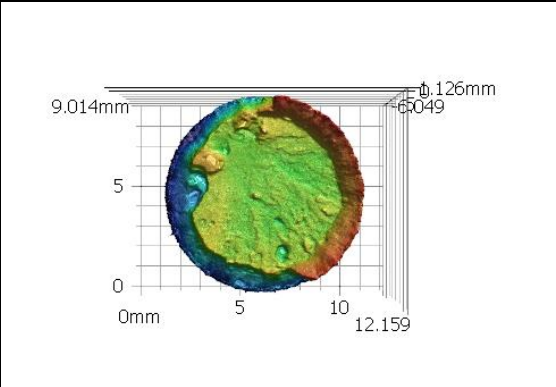
Specimen label	A-K6		Specimen label	A-K7	
End date	12/3/2022		End date	12/3/2022	
Area	63.61725	mm ²	Area	63.61725	mm ²
Length	25.4	mm	Length	25.4	mm
Maximum Load	62999.20312	N	Maximum Load	63663.57812	N
Tensile stress at Maximum Load	990.28491	MPa	Tensile stress at Maximum Load	1000.72827	MPa
Modulus (Automatic)	106791.8996	MPa	Modulus (Automatic)	106130.2216	MPa
Area under curve		Pa	Area under curve		Pa
% Elongation	0.15237	mm/mm	% Elongation	0.11314	mm/mm
Tensile stress at Break (Standard)	897.92548	MPa	Tensile stress at Break (Standard)	945.63501	MPa
Tensile stress at Yield (Offset 0.2 %)		MPa	Tensile stress at Yield (Offset 0.2 %)	930.27601	MPa
Tensile strain (Extension) at Maximum Load	0.0667	mm/mm	Tensile strain (Extension) at Maximum Load	0.06009	mm/mm

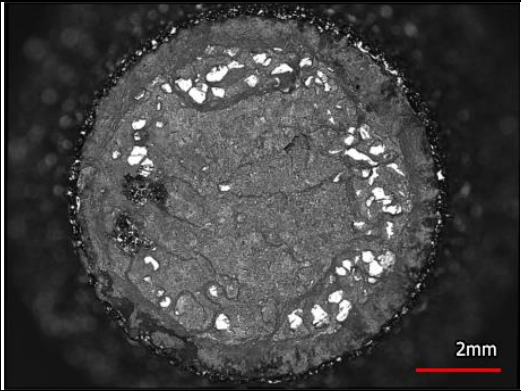
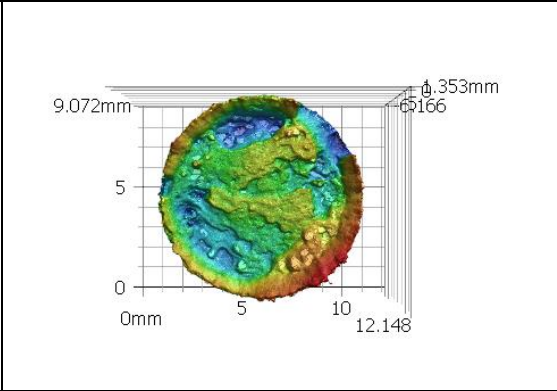
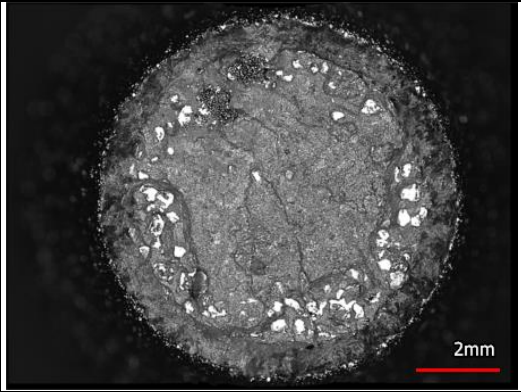
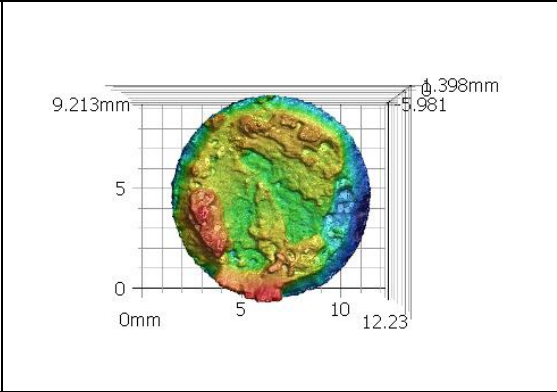
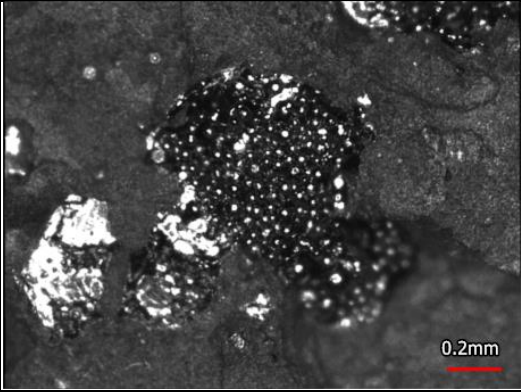
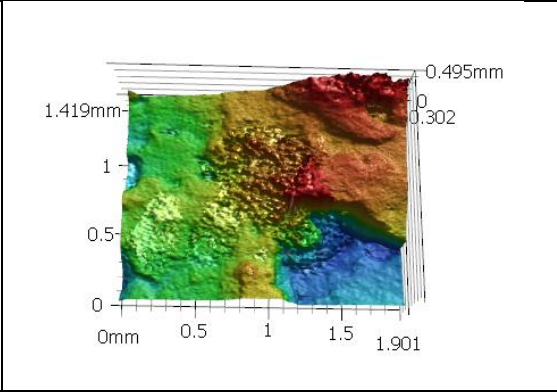
Optical image analysis of final tensile fracture surfaces

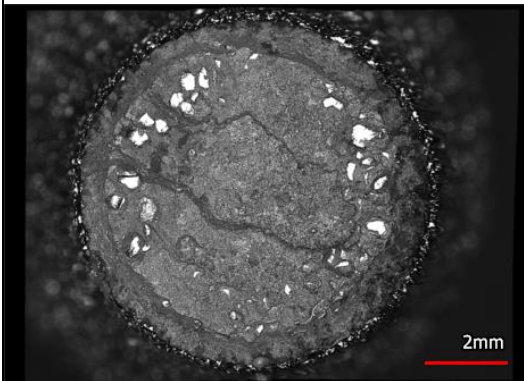
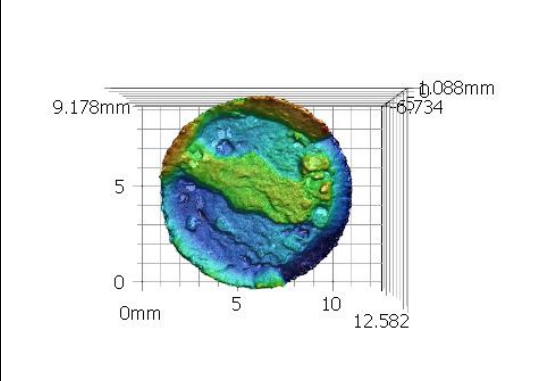
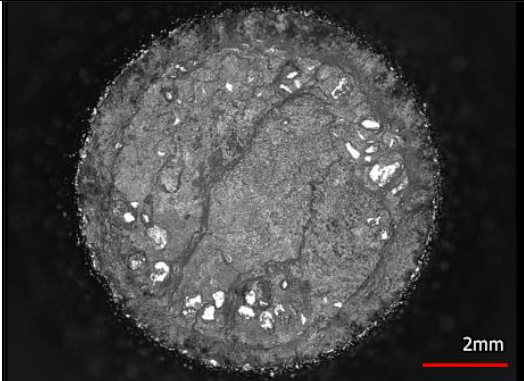
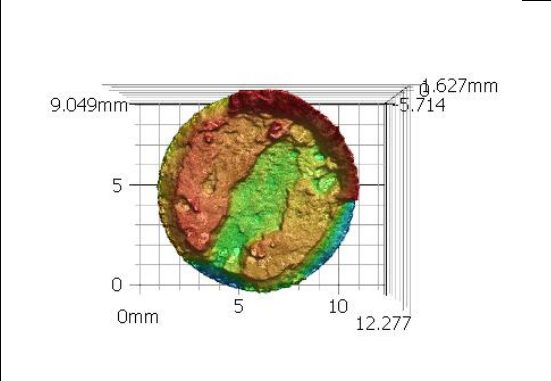
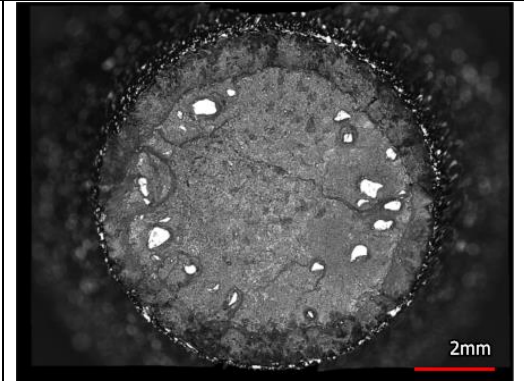
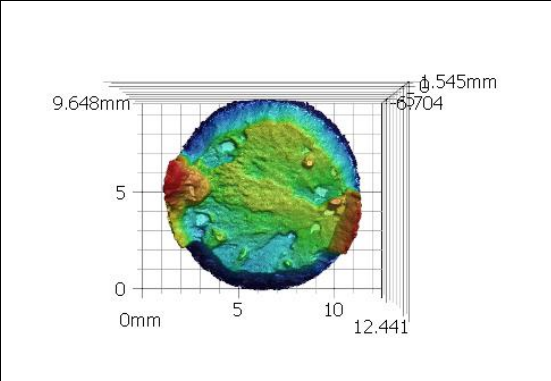
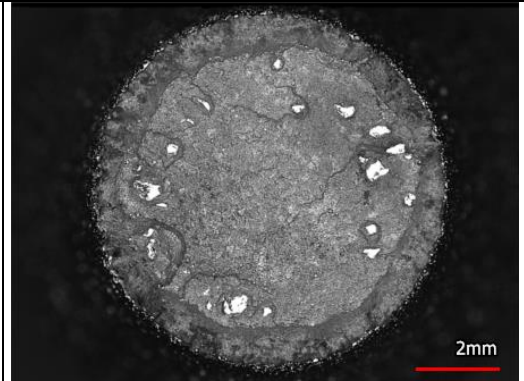
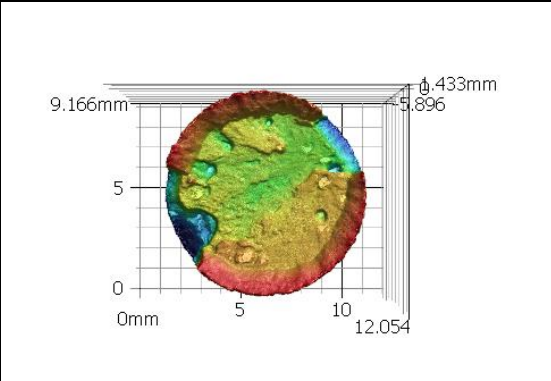
Measurement data name	Main image Optical	3D image
LG1-1		
LG1-2		

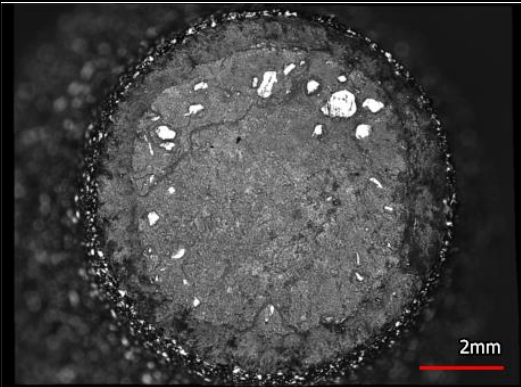
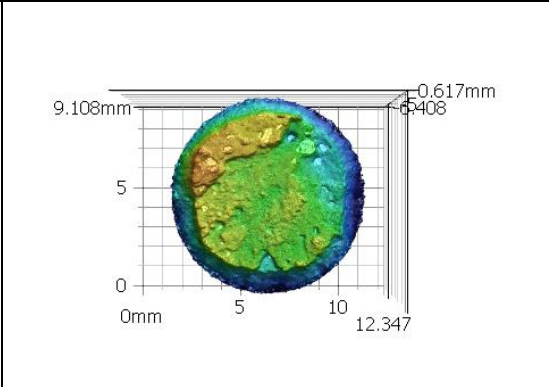
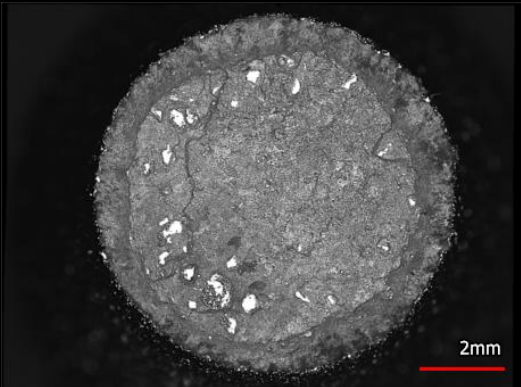
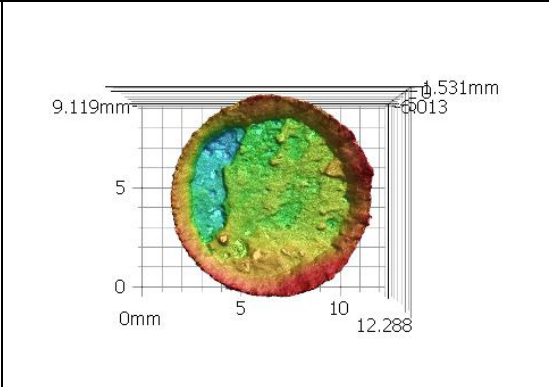
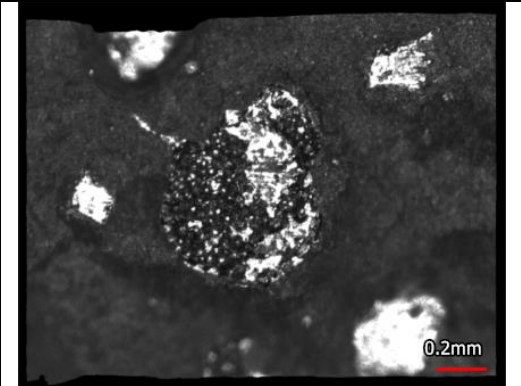
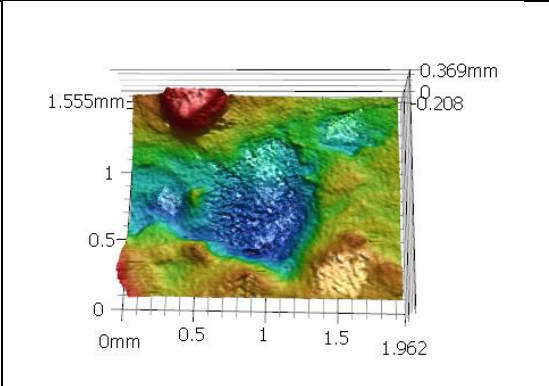
Measurement data name	Main image Optical	3D image
LG2-1		
LG2-2		
LG3-1		
LG3-2		

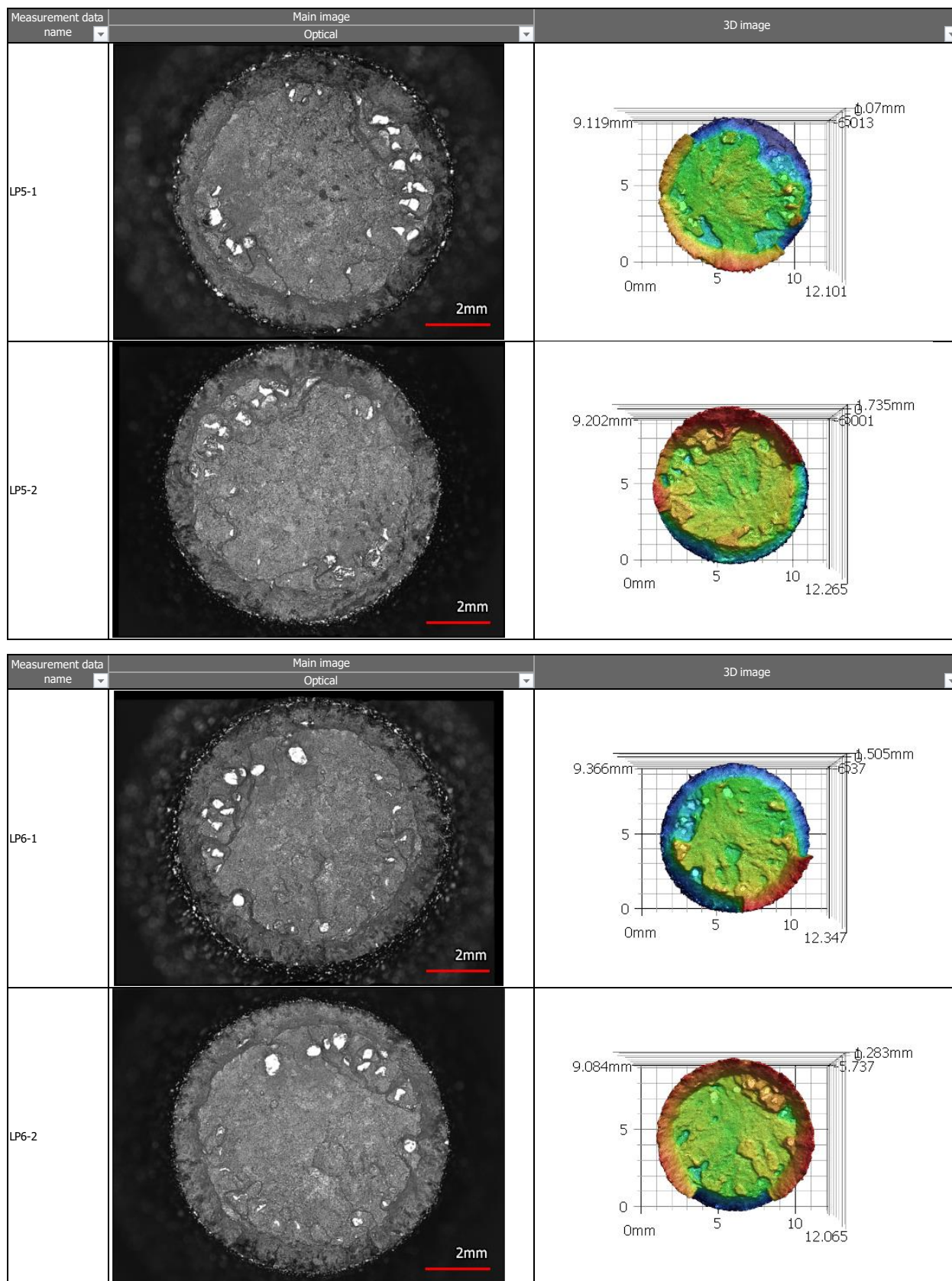
Measurement data name	Main image Optical	3D image
LG4-1		
LG4-2		
Measurement data name	Main image Optical	3D image
LG5-1		
LG5-2		

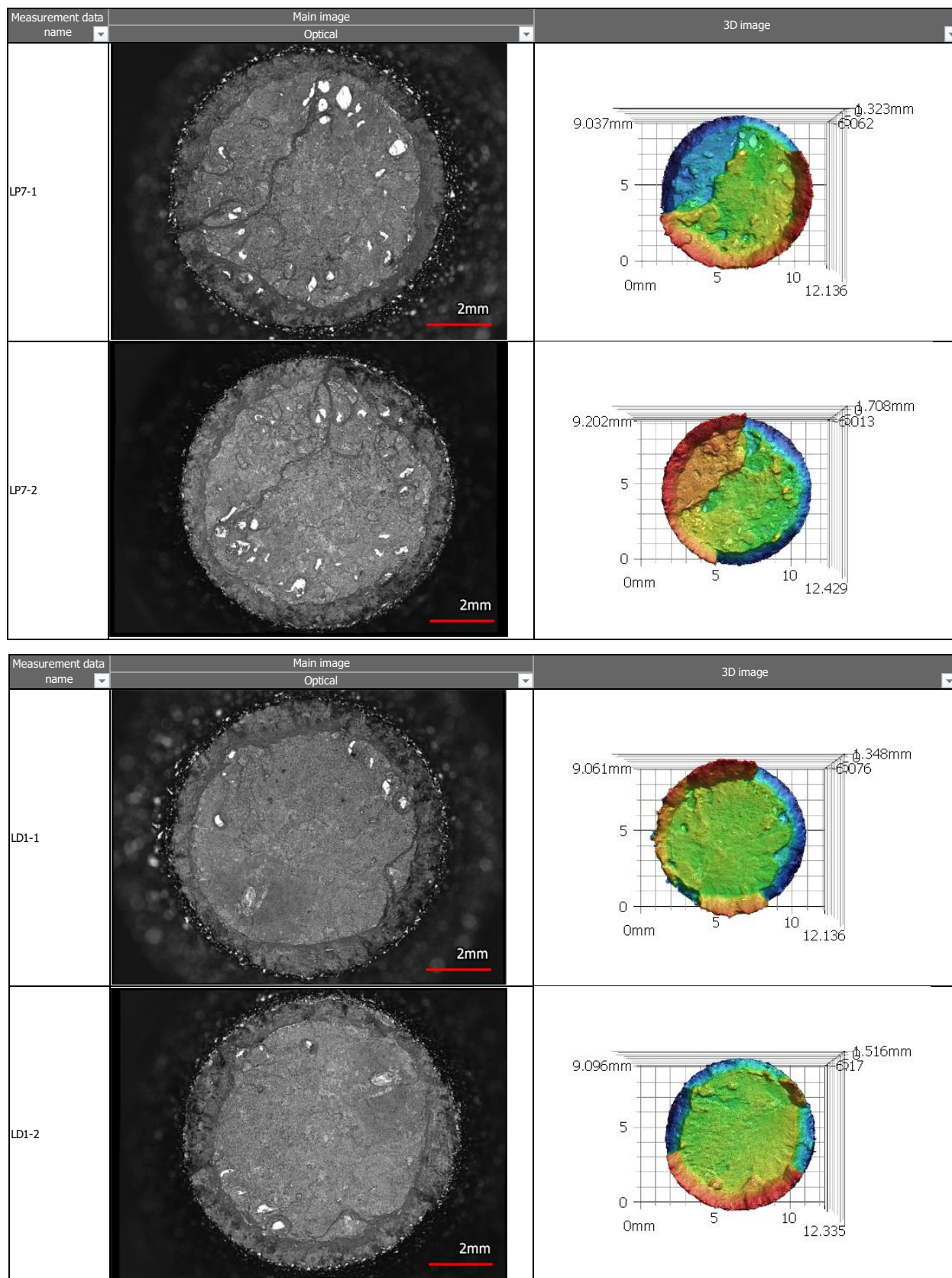
Measurement data name	Main image Optical	3D image
LG6-1		
LG6-2		
Measurement data name	Main image Optical	3D image
LG7-1		
LG7-2		

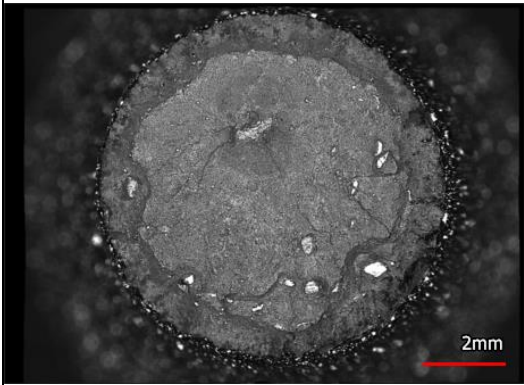
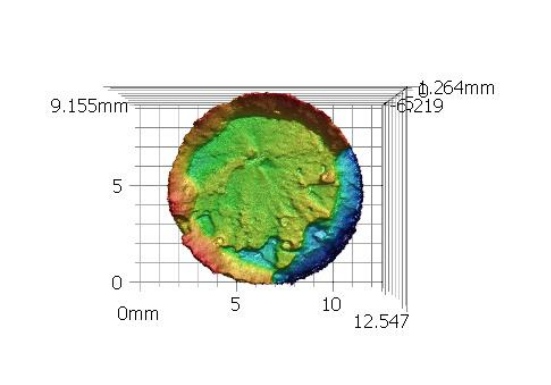
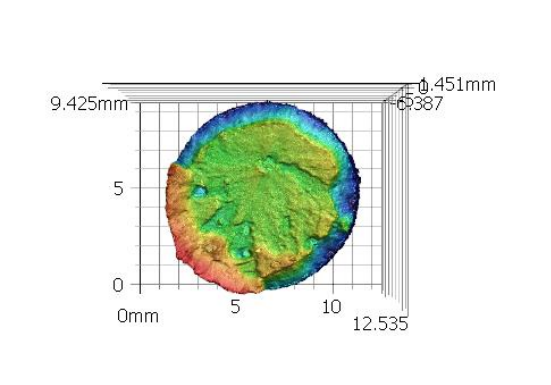
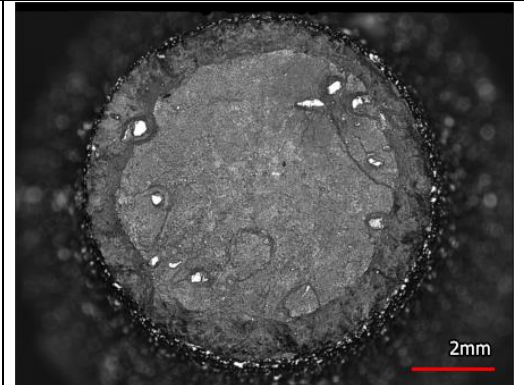
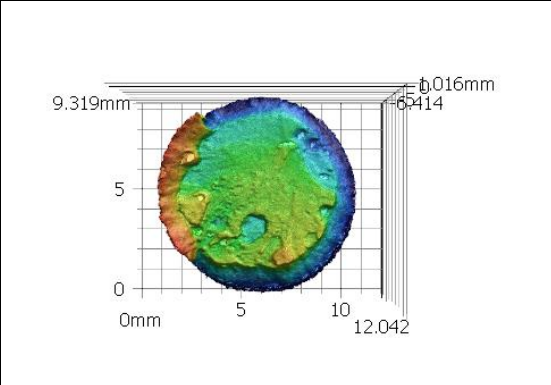
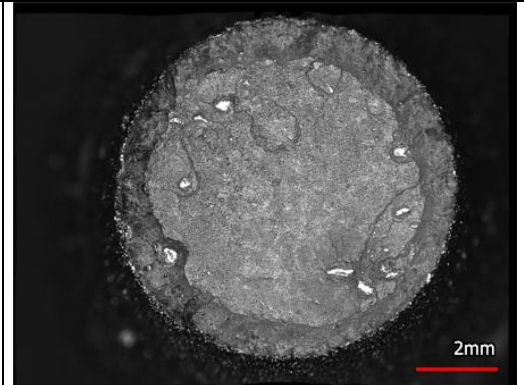
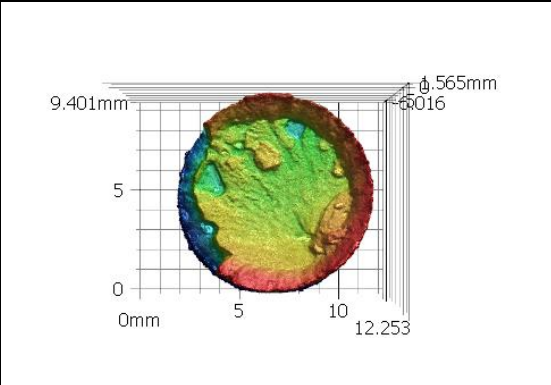
Measurement data name	Main image Optical	3D image
LP1-1		
LP1-2		
LP1-2-powder		

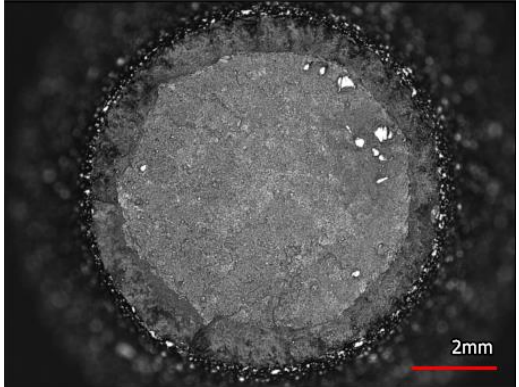
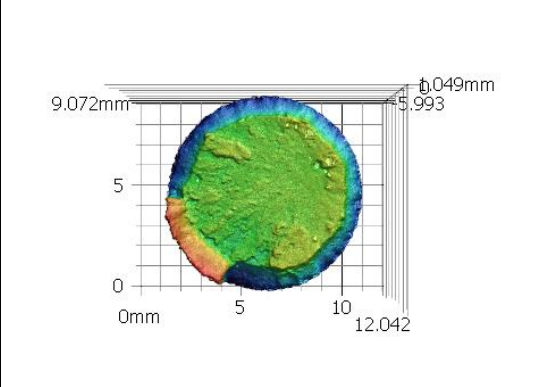
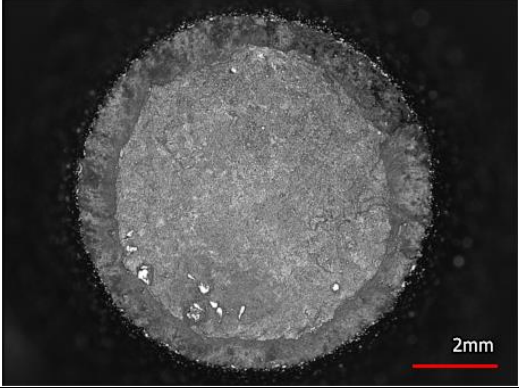
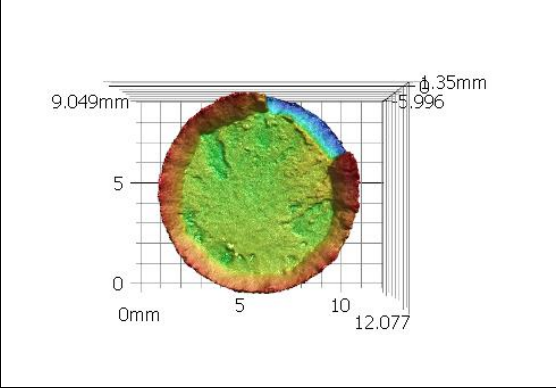
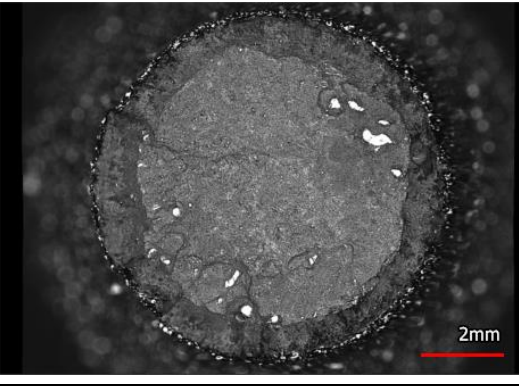
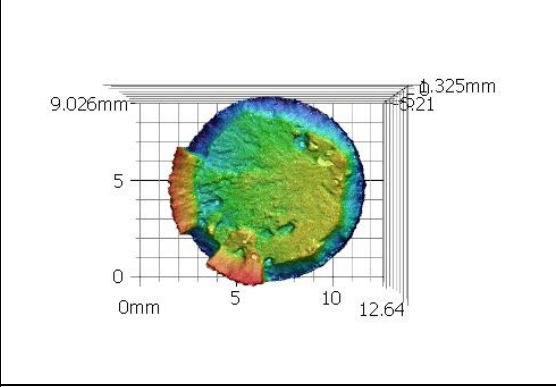
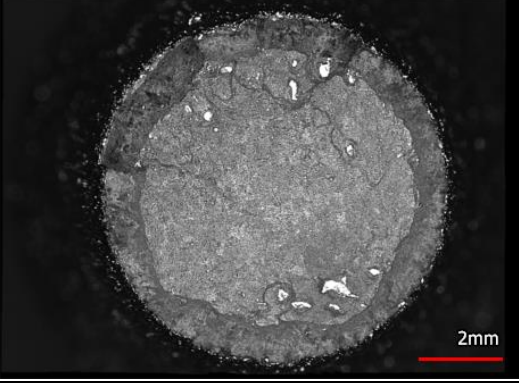
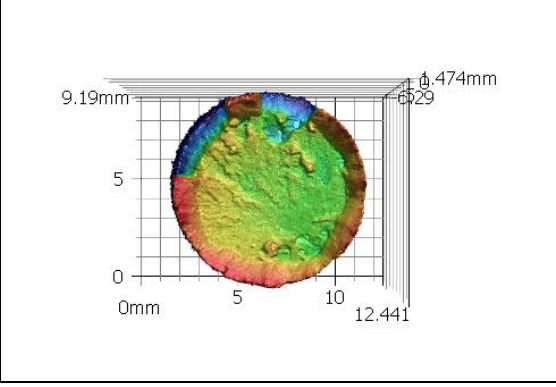
Measurement data name	Main image Optical	3D image
LP2-1		
LP2-2		
LP3-1		
LP3-2		

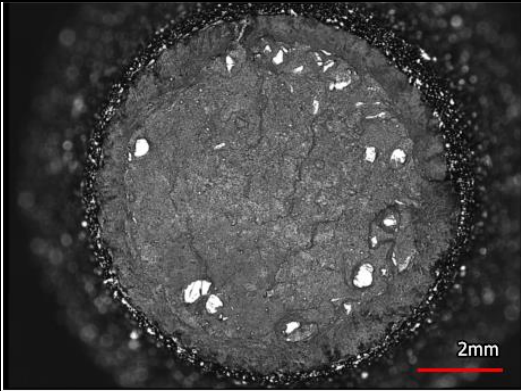
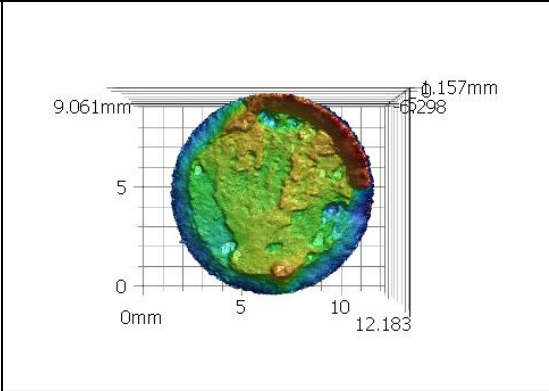
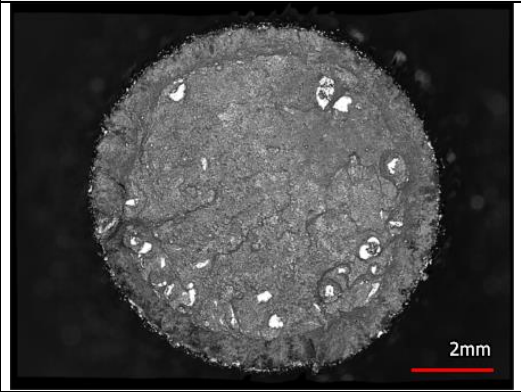
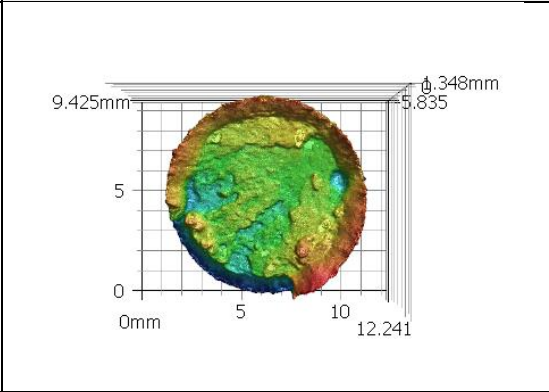
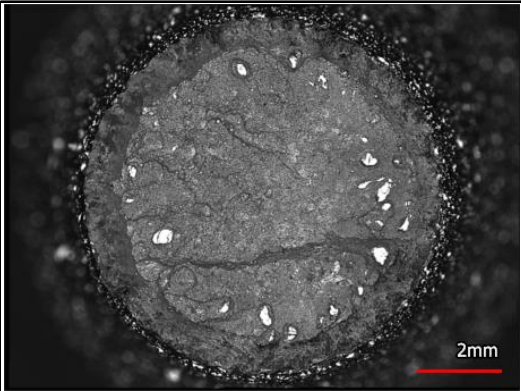
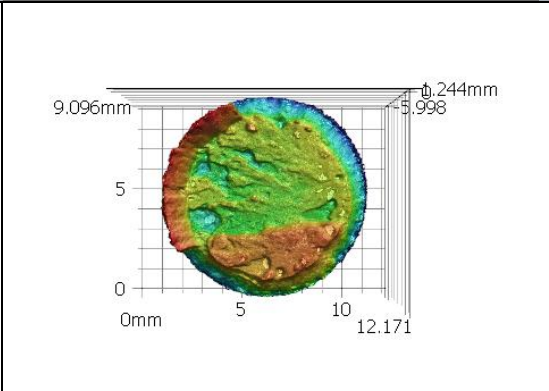
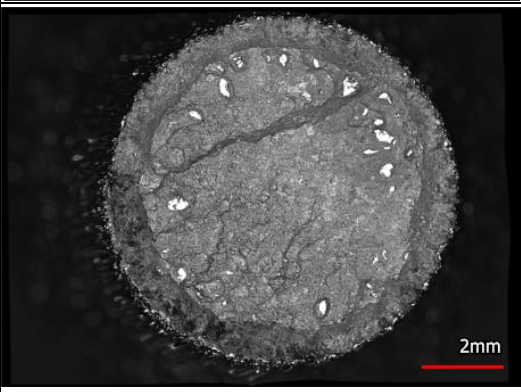
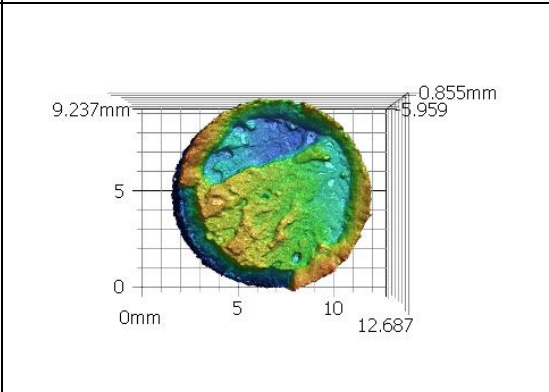
Measurement data name	Main image Optical	3D image
LP4-1		
LP4-2		
LP4-2-powder		

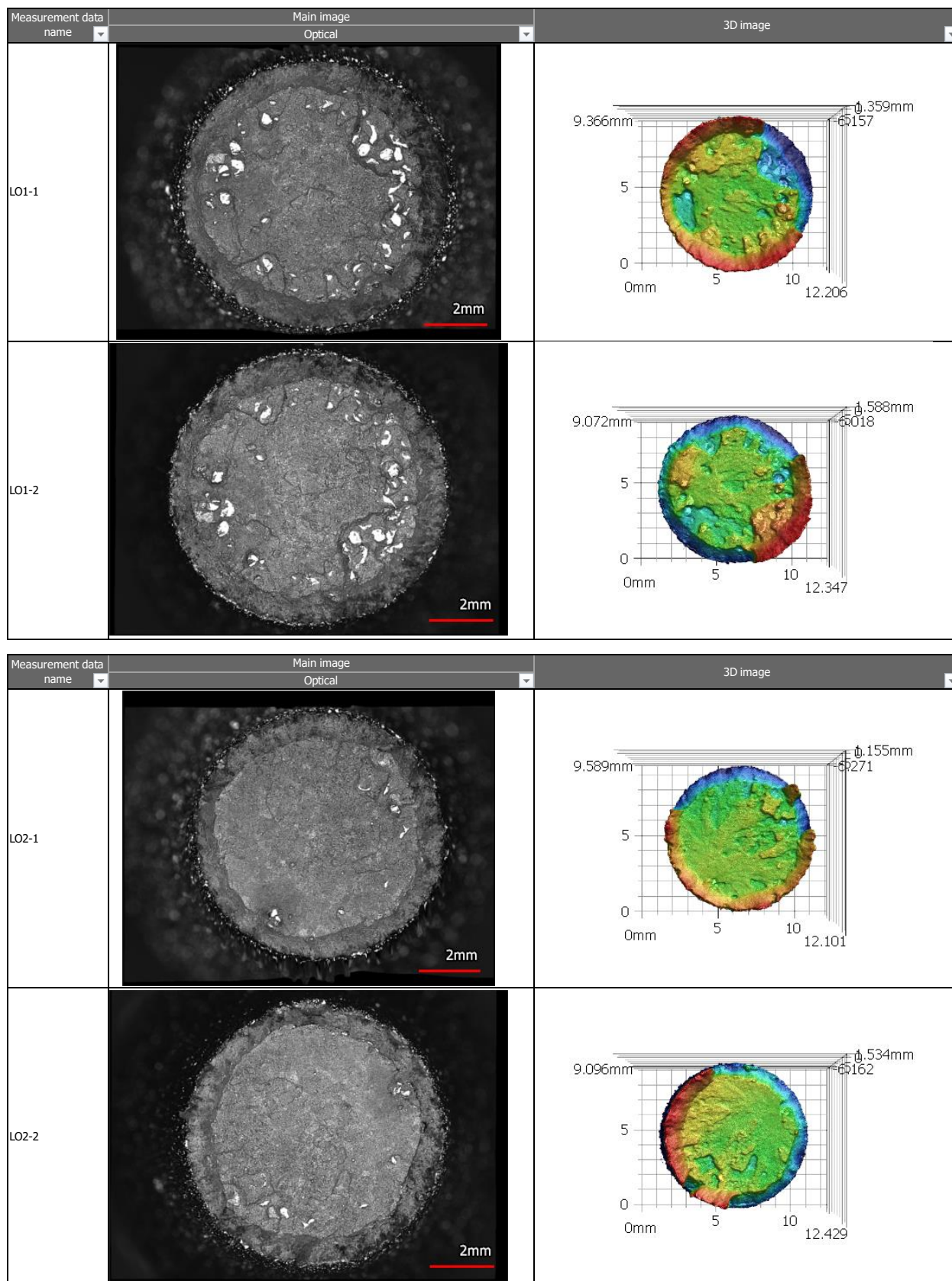


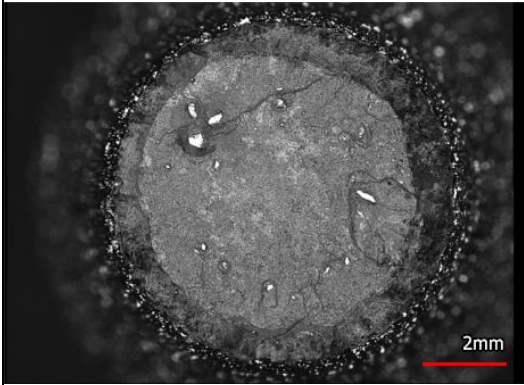
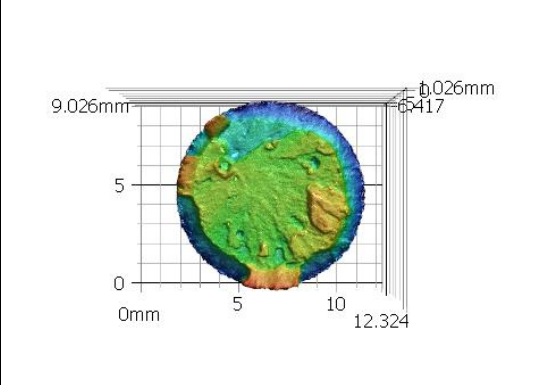
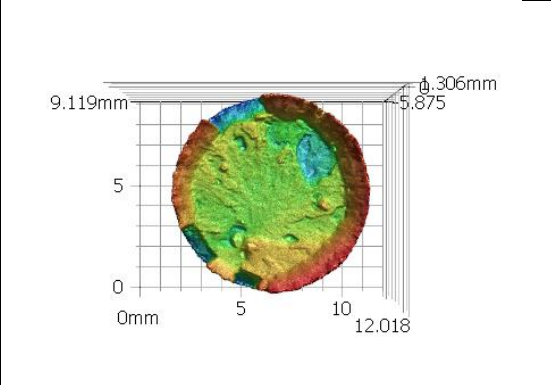
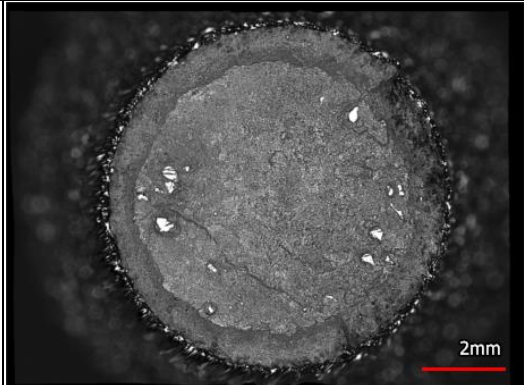
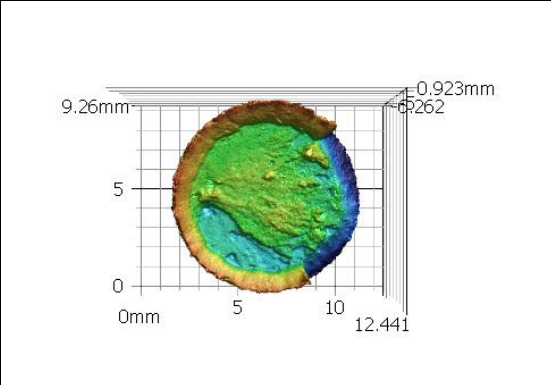
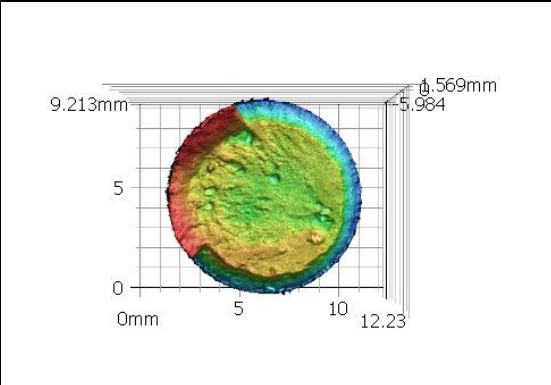


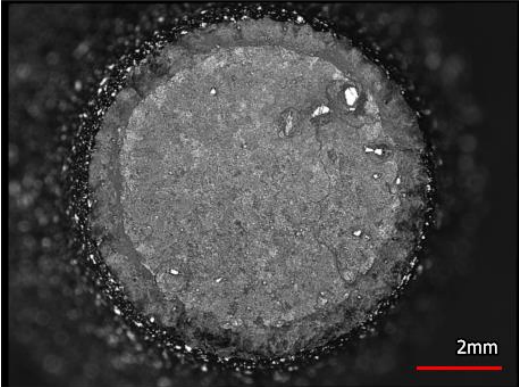
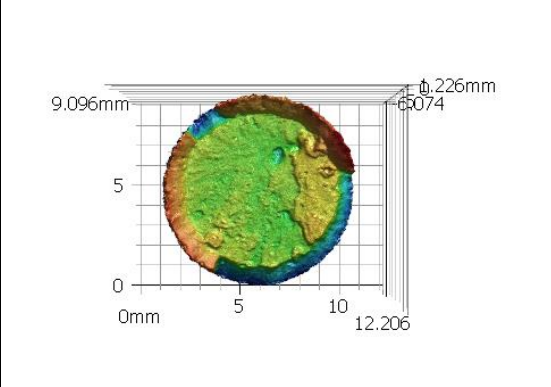
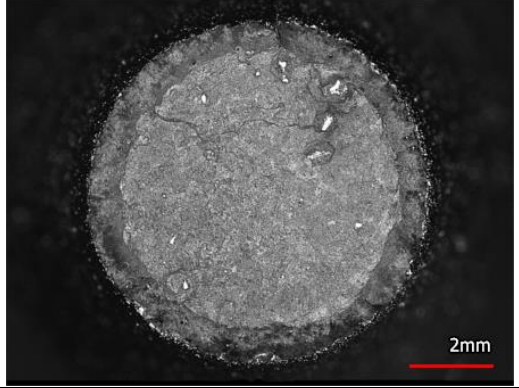
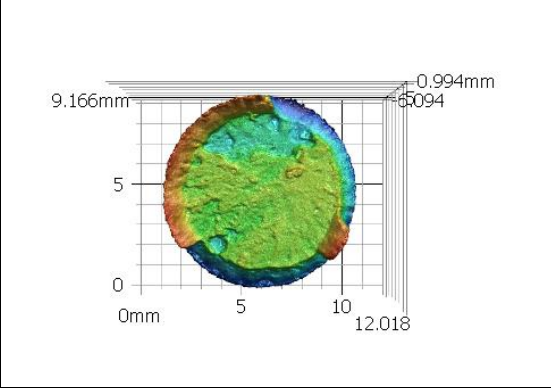
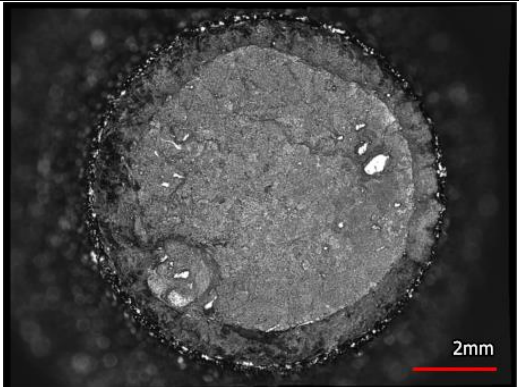
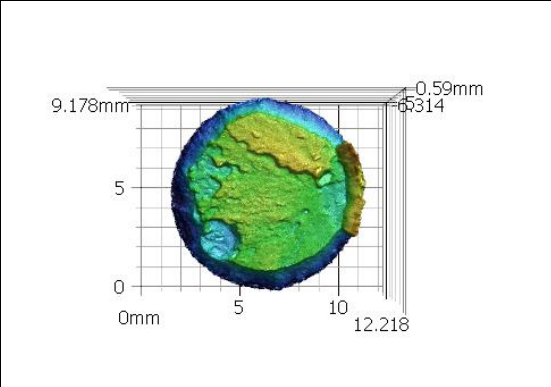
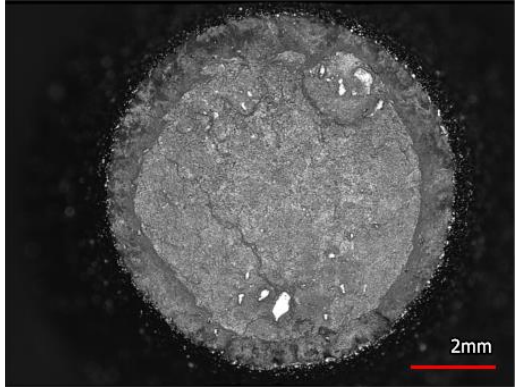
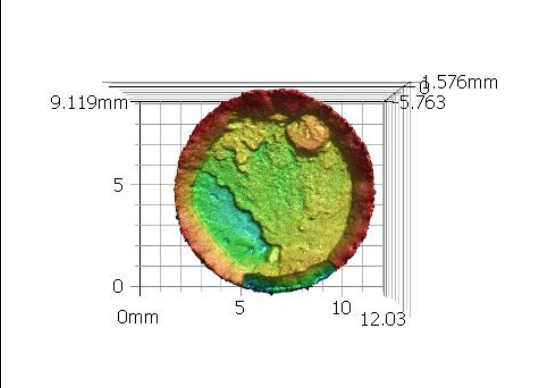
Measurement data name	Main image Optical	3D image
LD2-1		
LD2-2		
Measurement data name	Main image Optical	3D image
LD3-1		
LD3-2		

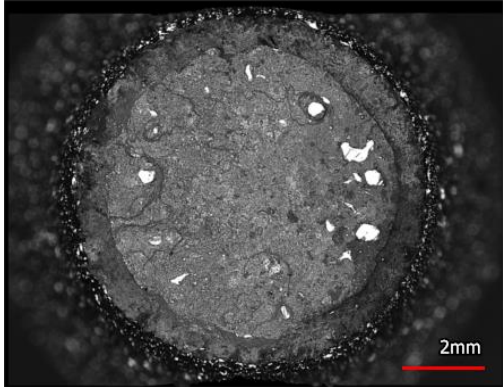
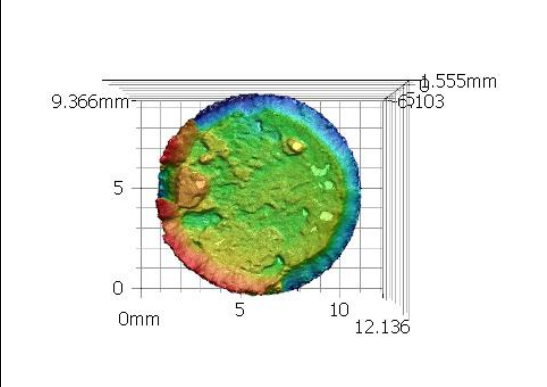
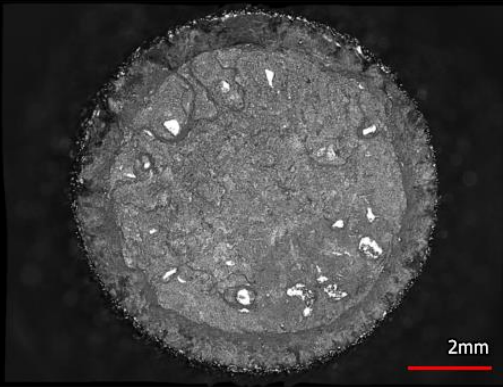
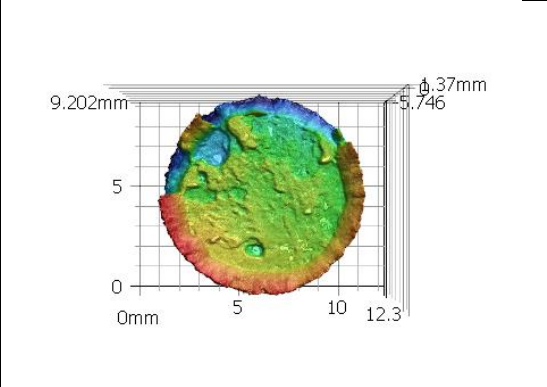
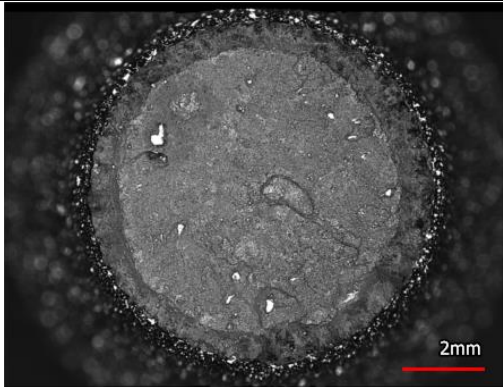
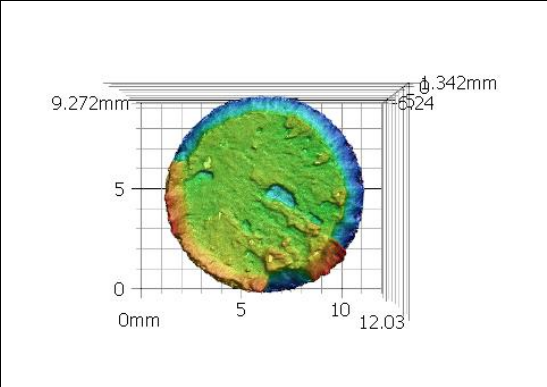
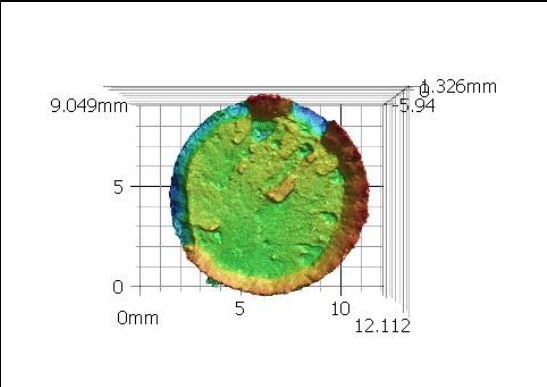
Measurement data name	Main image Optical	3D image
LD4-1		
LD4-2		
LD5-1		
LD5-2		

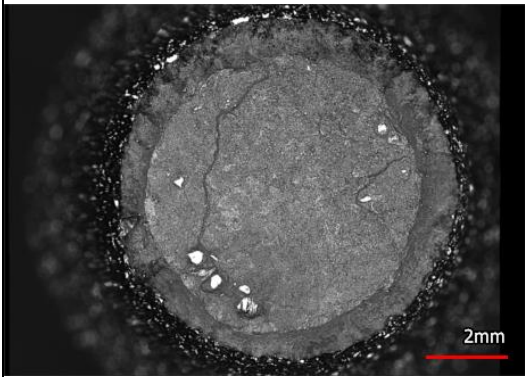
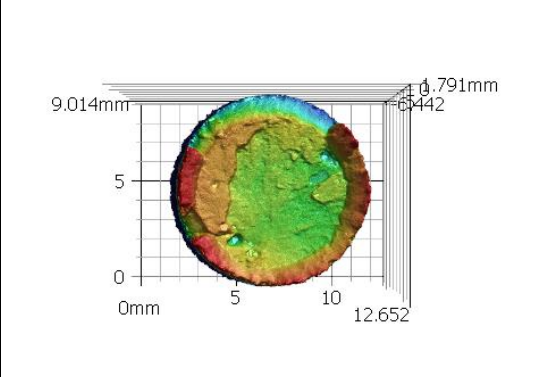
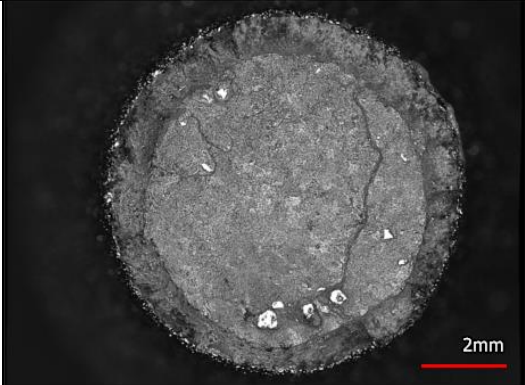
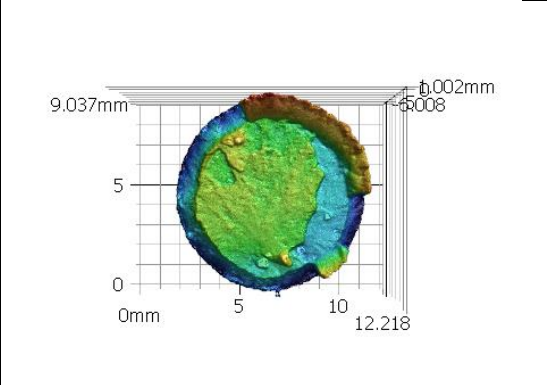
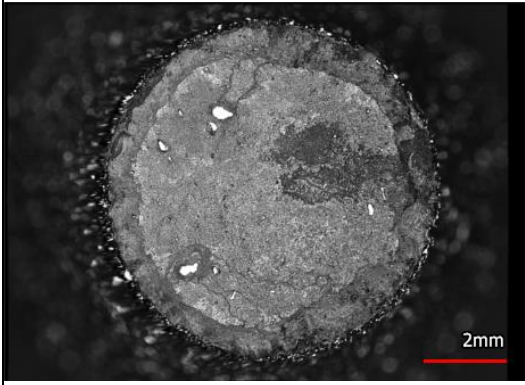
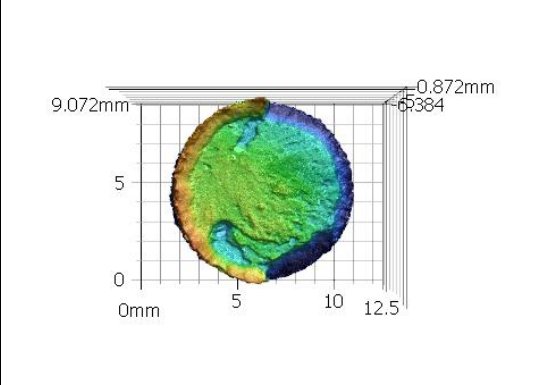
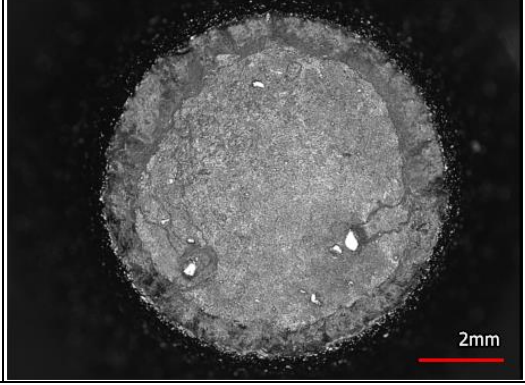
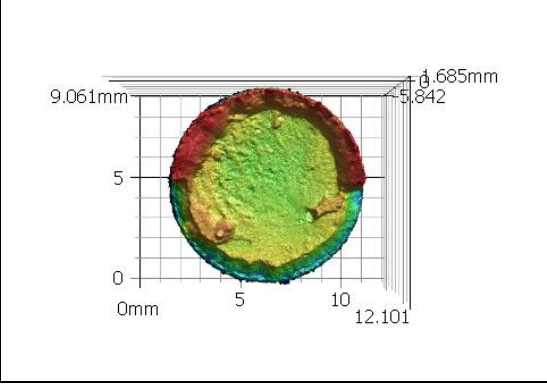
Measurement data name	Main image Optical	3D image
LD6-1		
LD6-2		
LD7-1		
LD7-2		

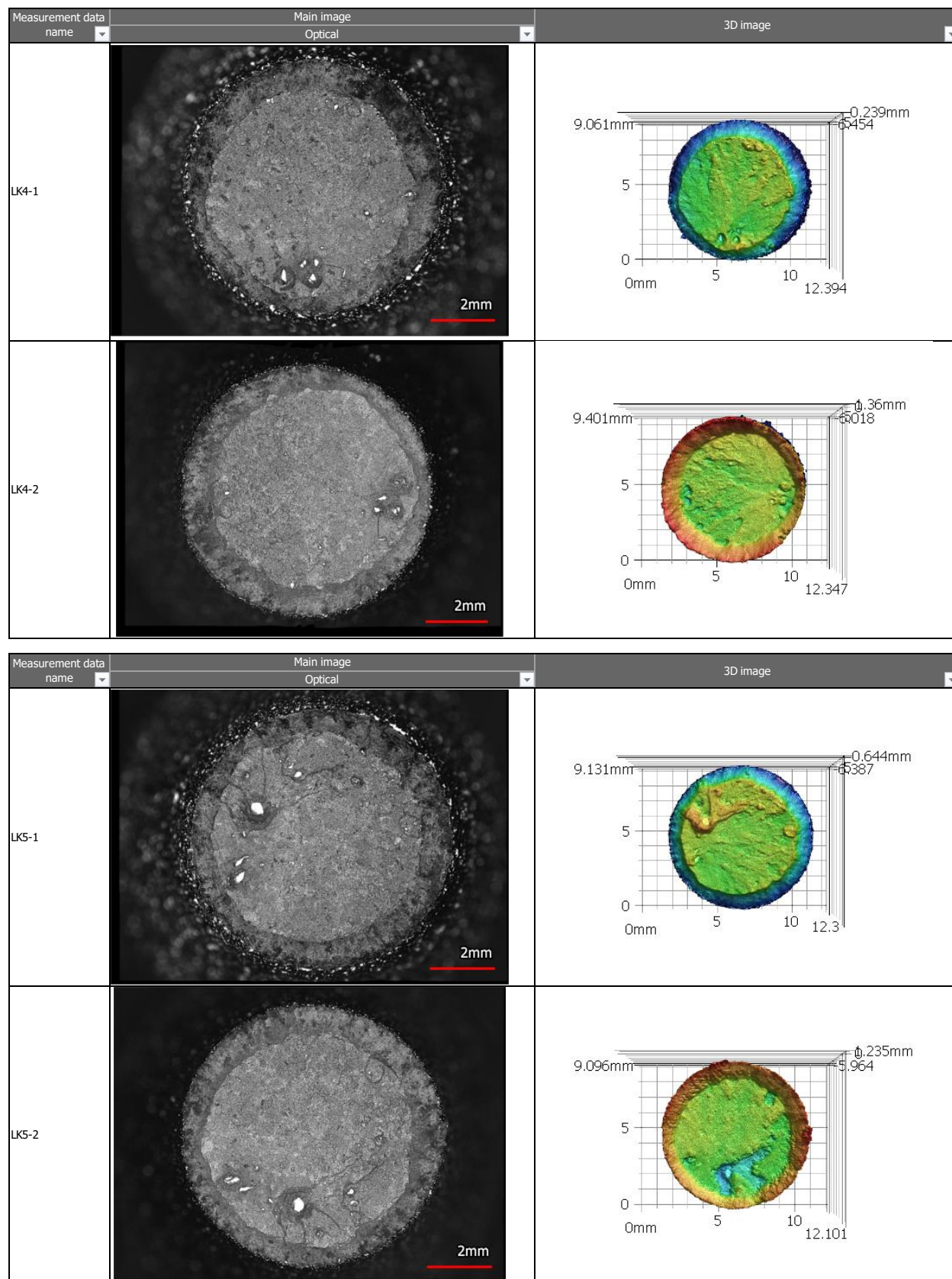


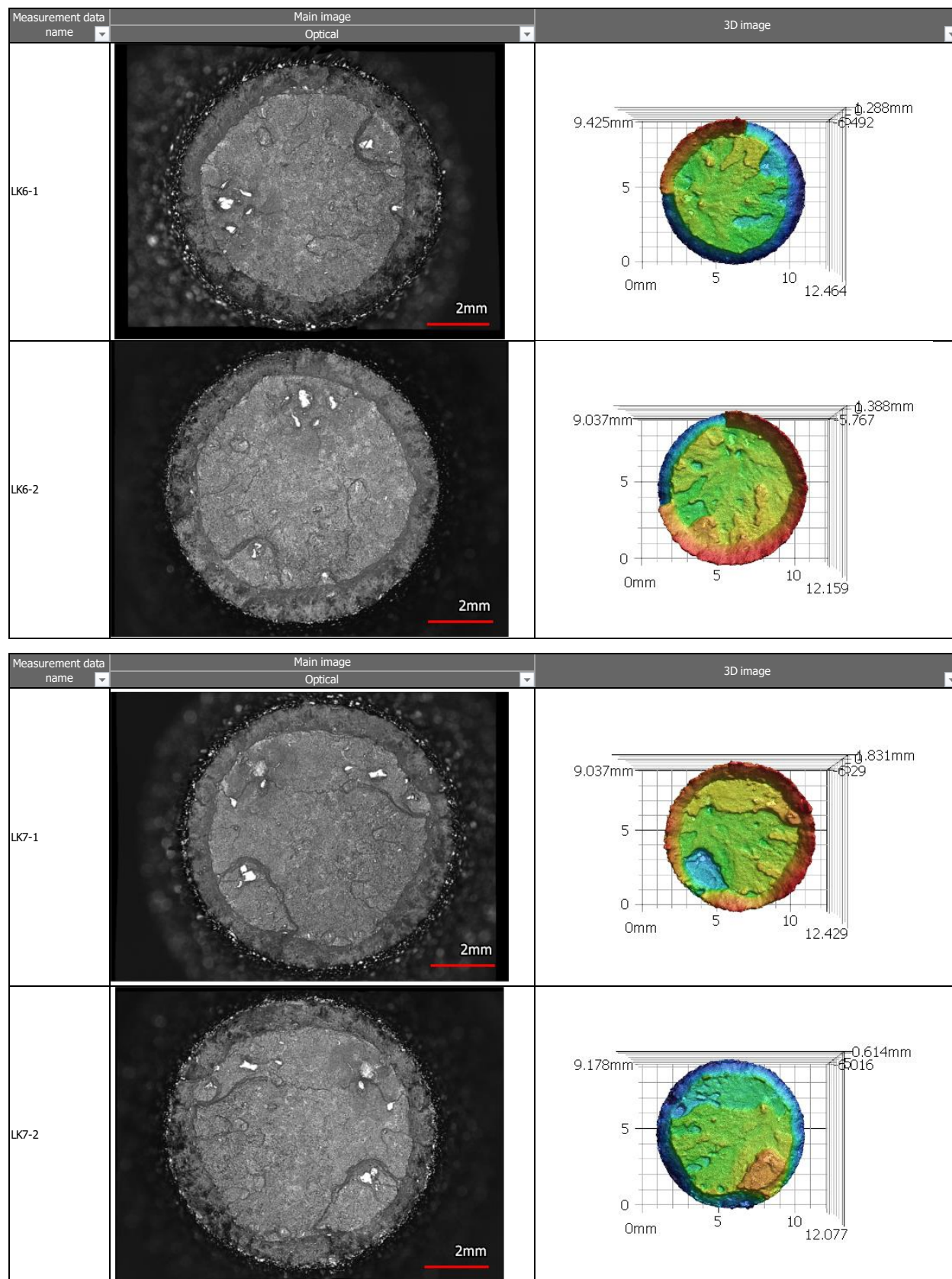
Measurement data name	Main image Optical	3D image
L03-1		 9.026mm 5 0 0mm 5 10 12.324 0.026mm 5.417
L03-2		 9.119mm 5 0 0mm 5 10 12.018 0.306mm 5.875
Measurement data name	Main image Optical	3D image
L04-1		 9.26mm 5 0 0mm 5 10 12.441 0.923mm 5.262
L04-2		 9.213mm 5 0 0mm 5 10 12.23 0.569mm 5.984

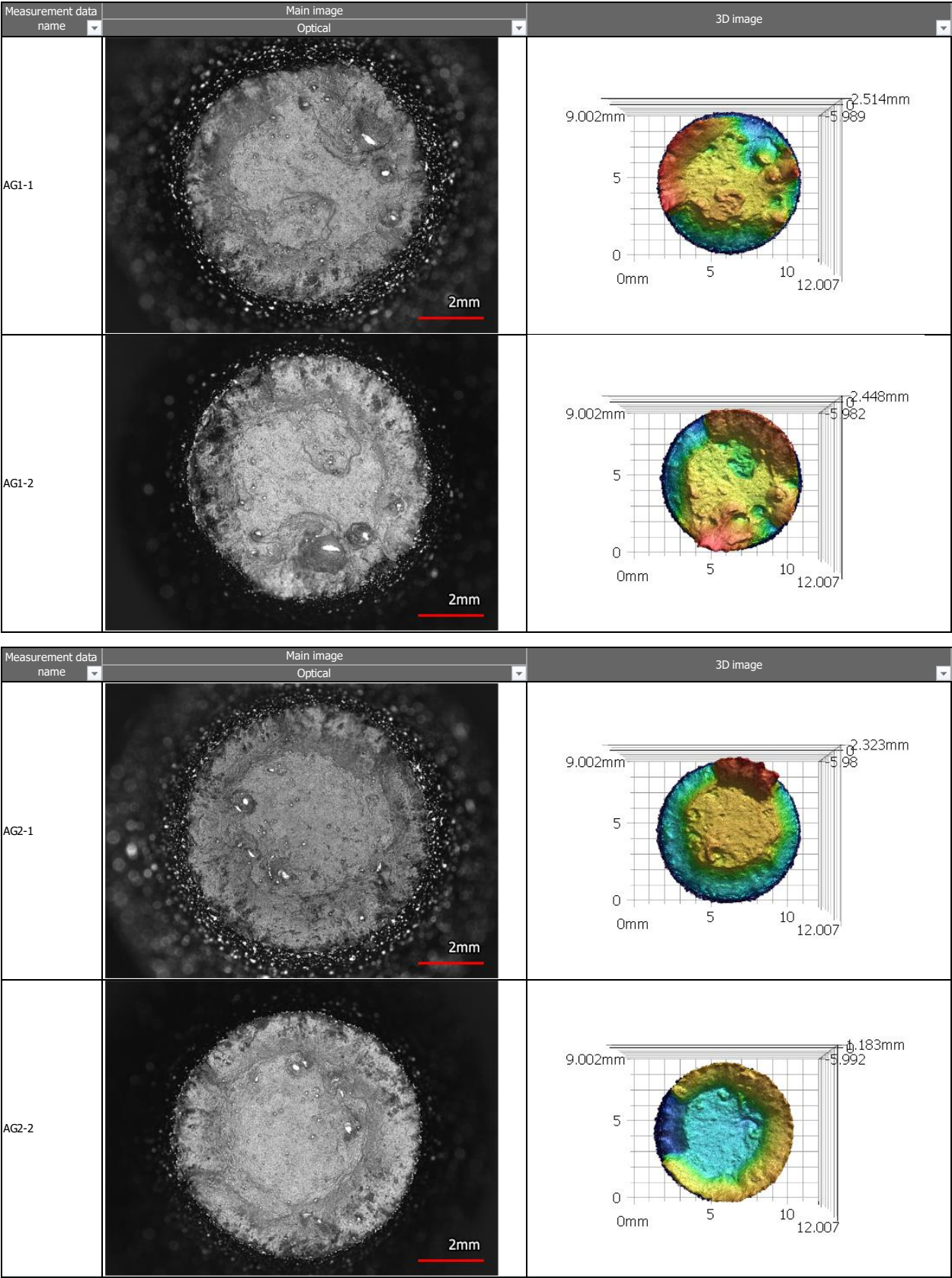
Measurement data name	Main image Optical	3D image
L05-1		
L05-2		
L06-1		
L06-2		

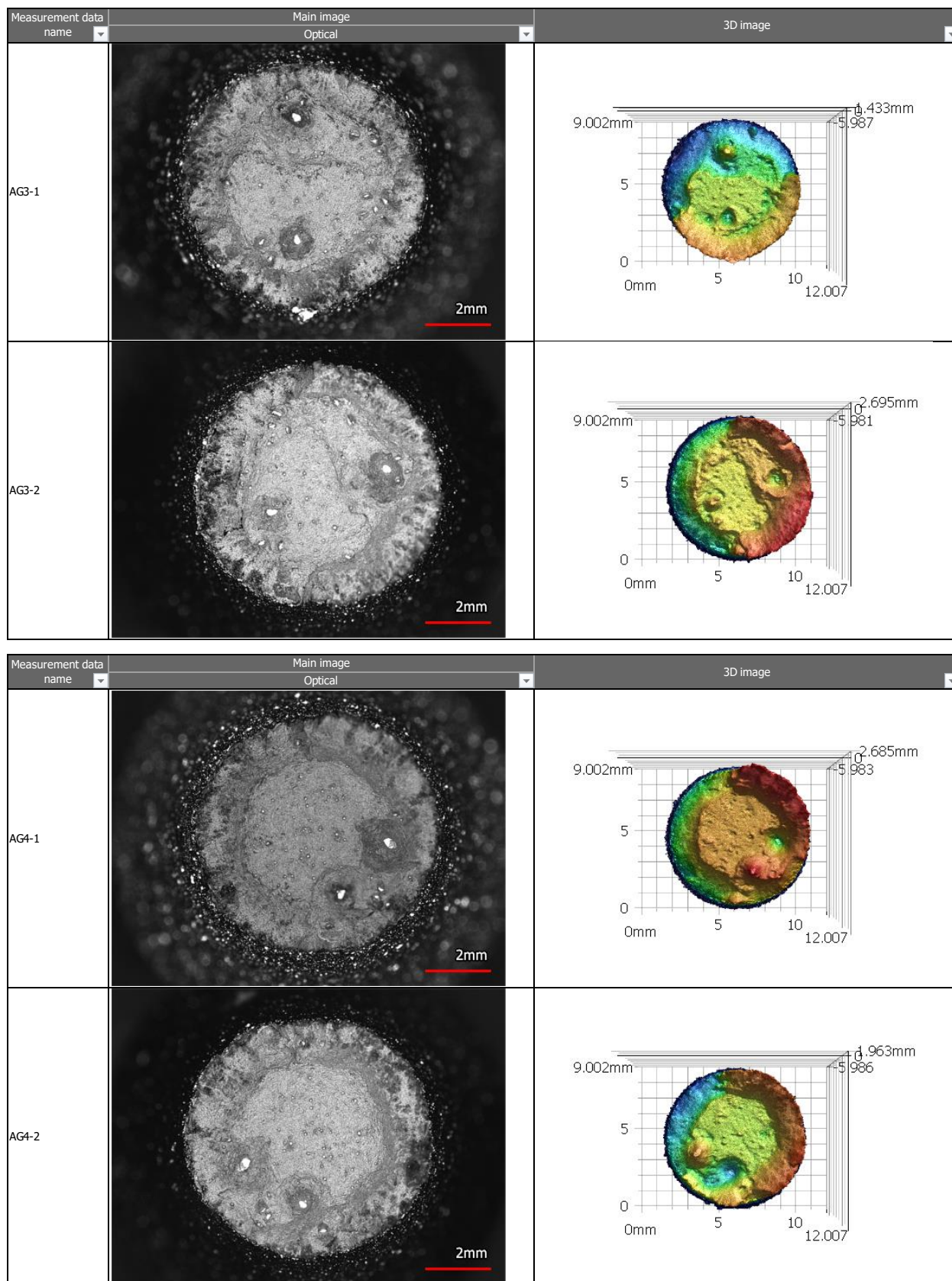
Measurement data name	Main image Optical	3D image
L07-1		
L07-2		
LK1-1		
LK1-2		

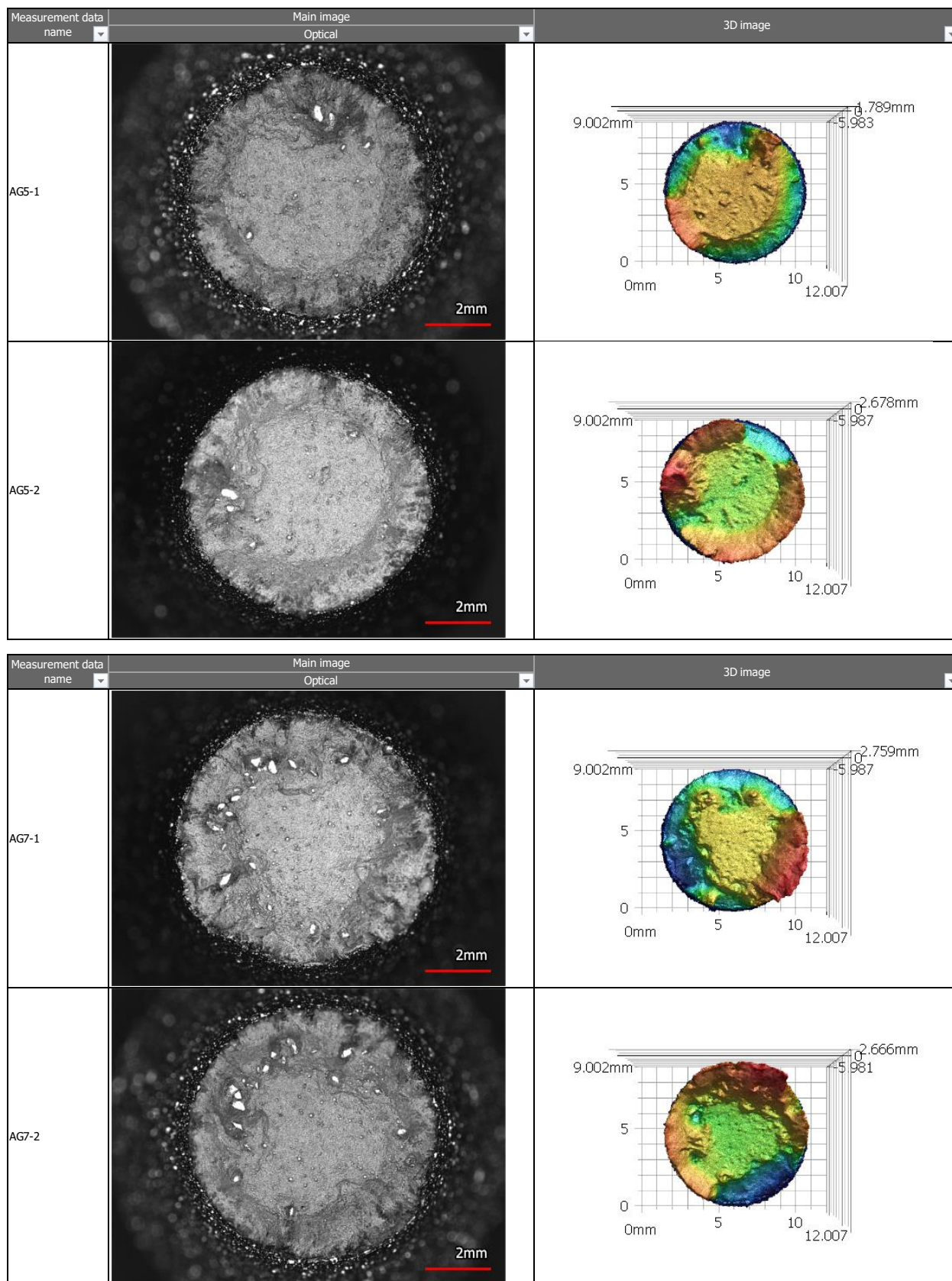
Measurement data name	Main image Optical	3D image
LK2-1		
LK2-2		
Measurement data name	Main image Optical	3D image
LK3-1		
LK3-2		

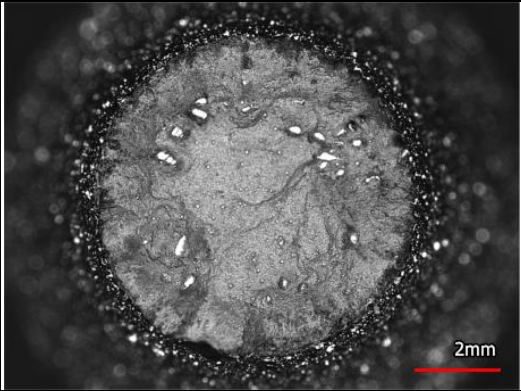
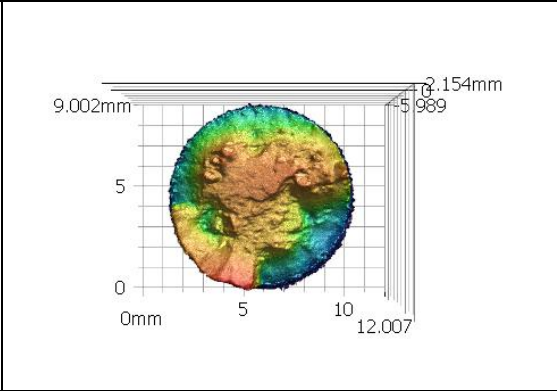
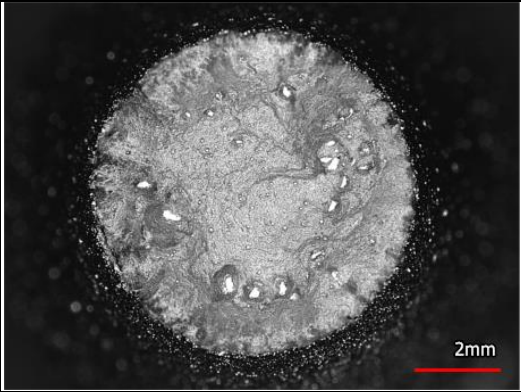
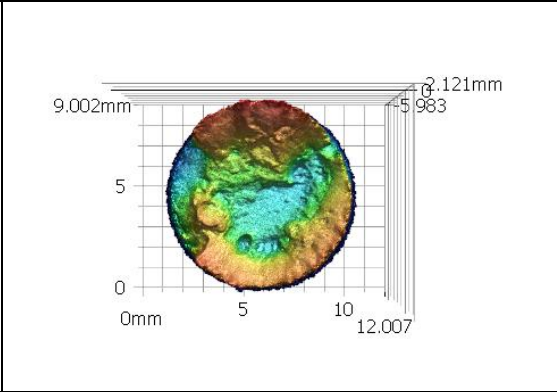
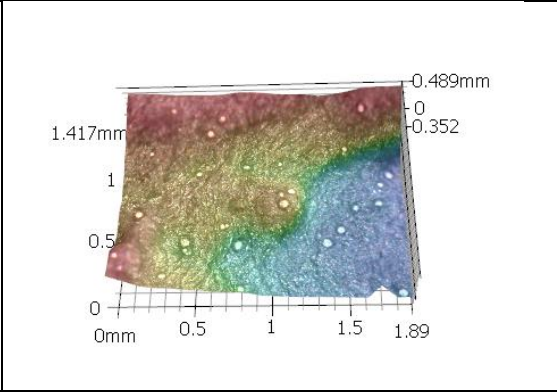


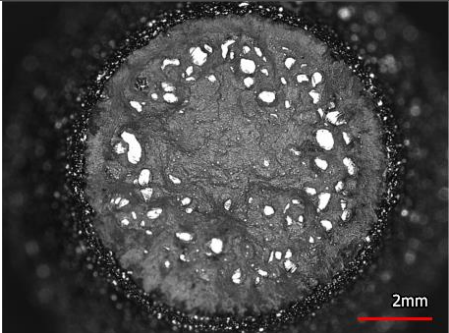
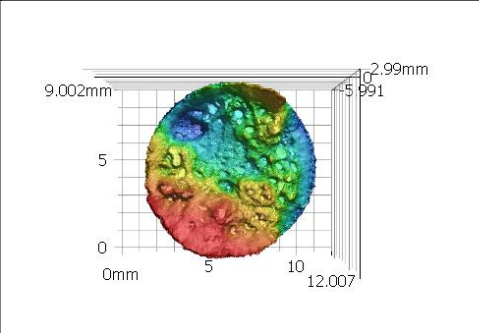
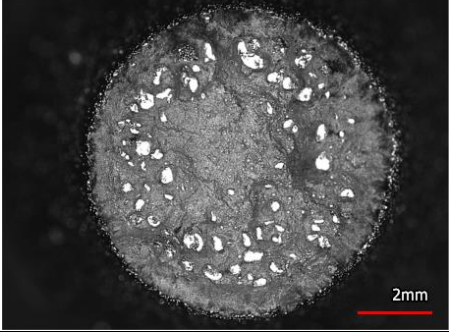
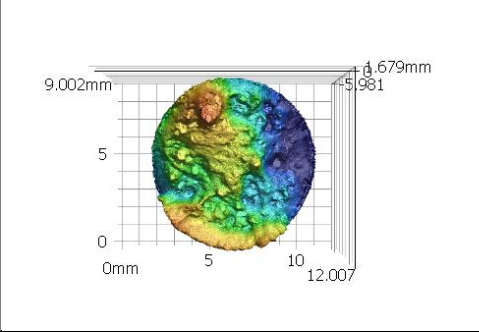
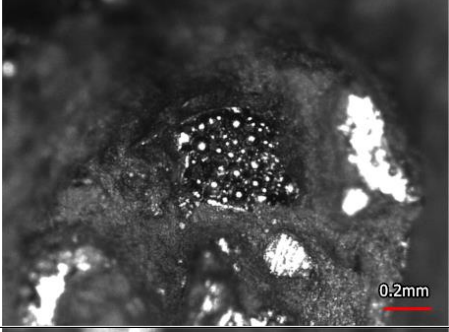
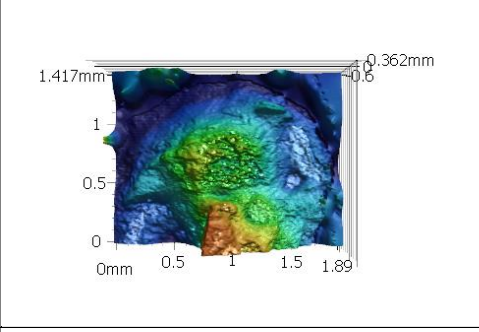
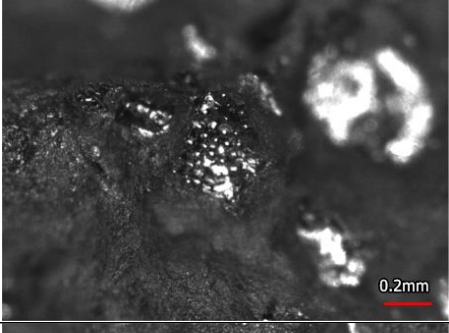
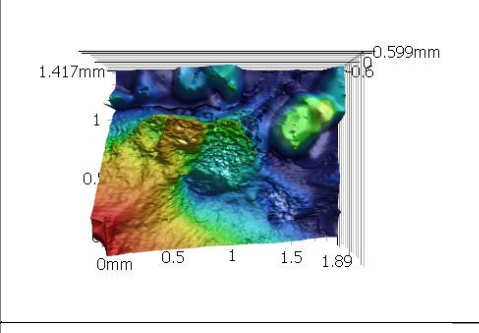
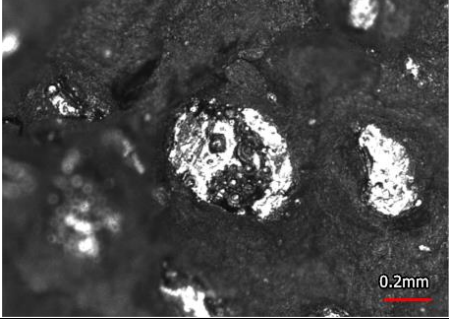
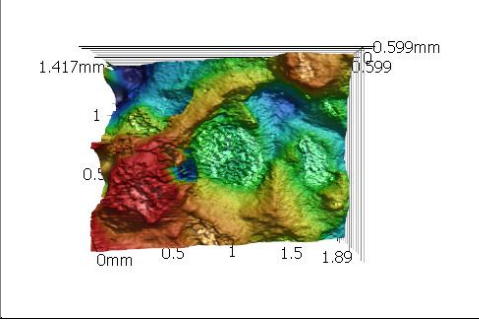


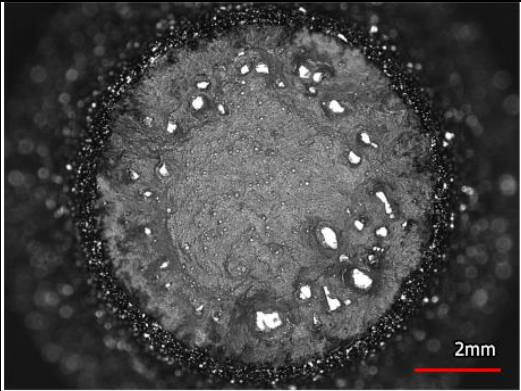
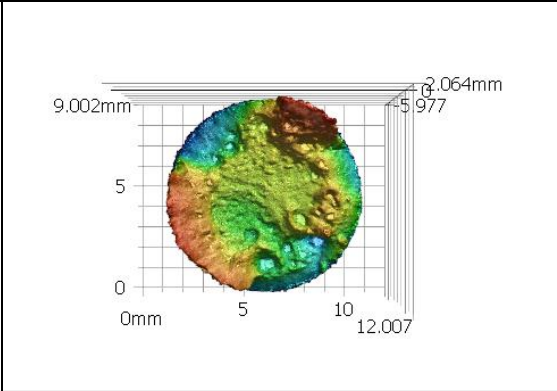
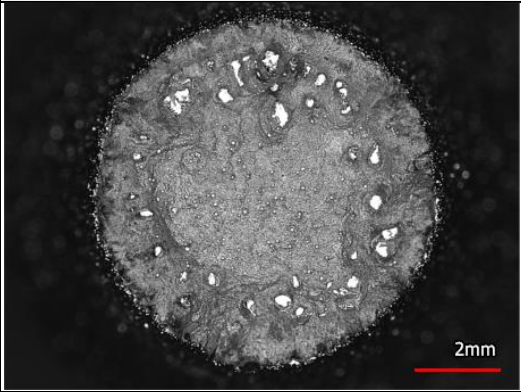
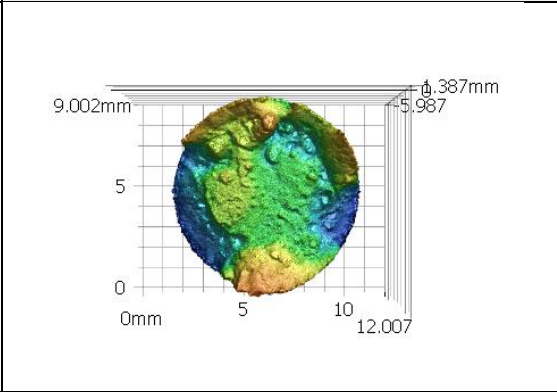
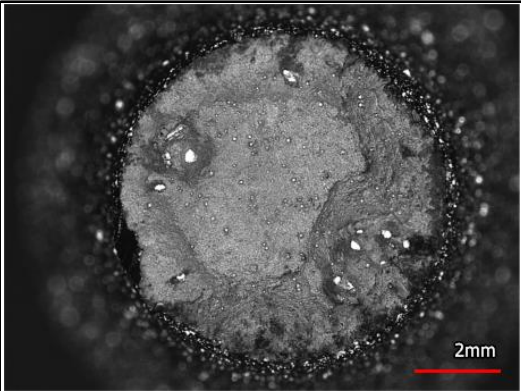
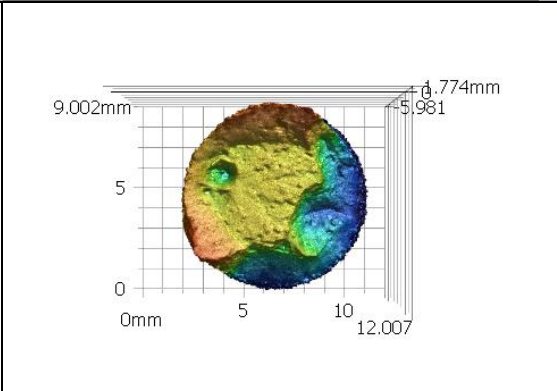
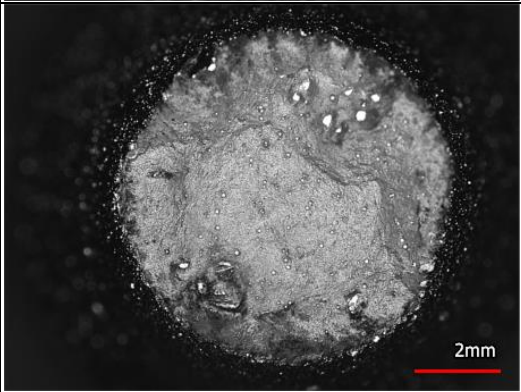
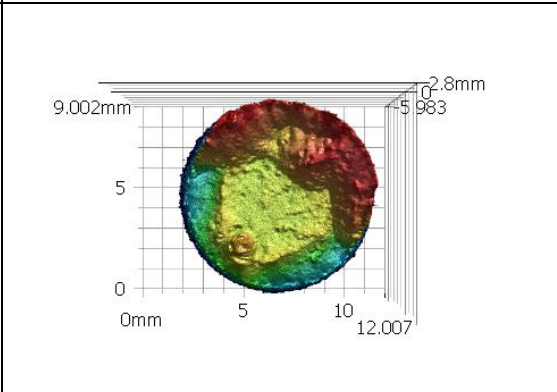


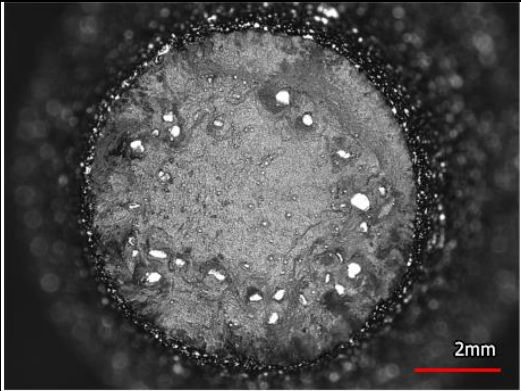
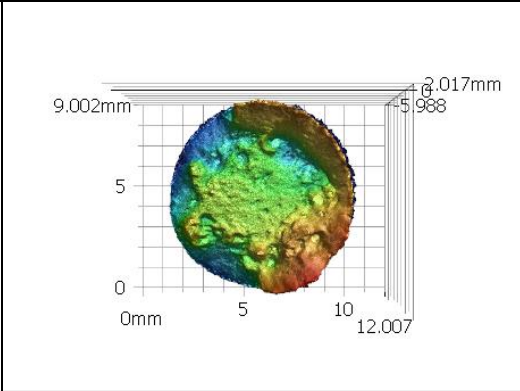
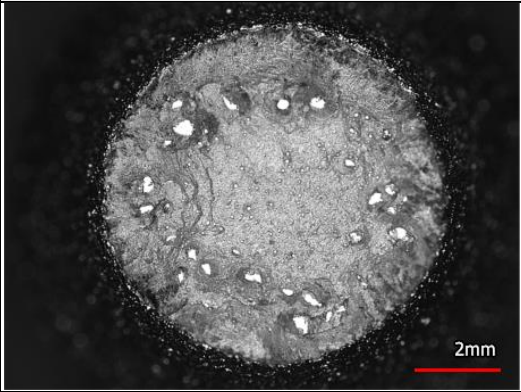
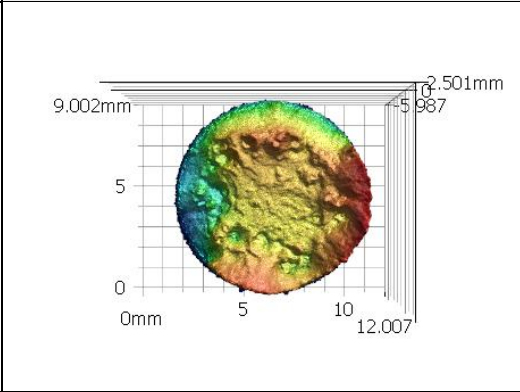
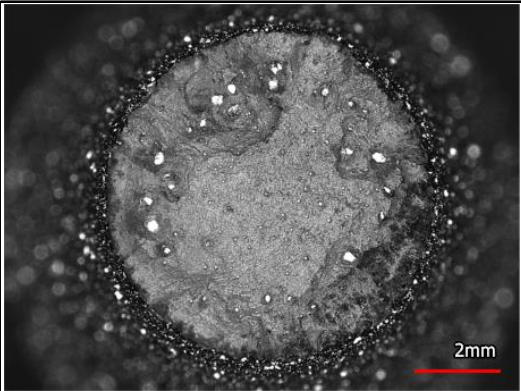
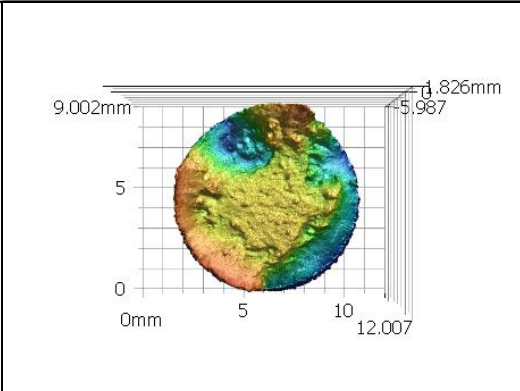
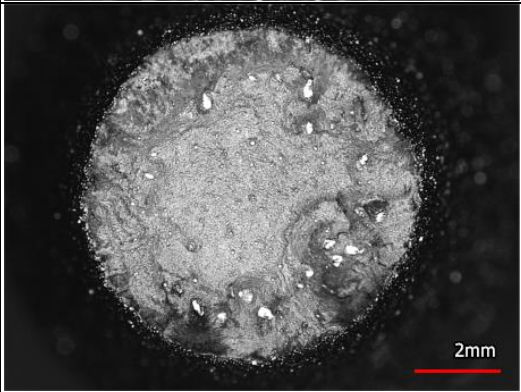
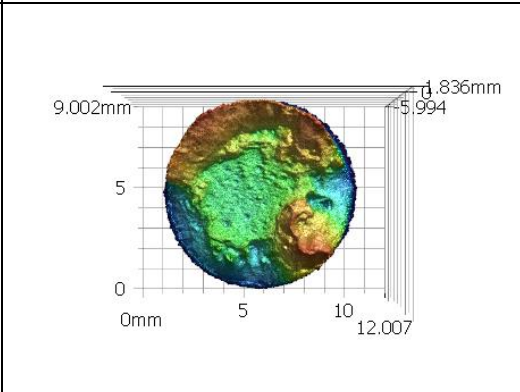


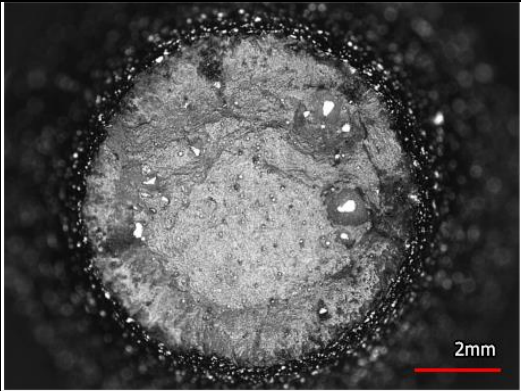
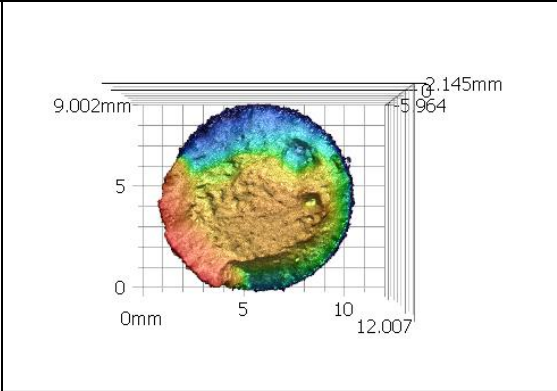
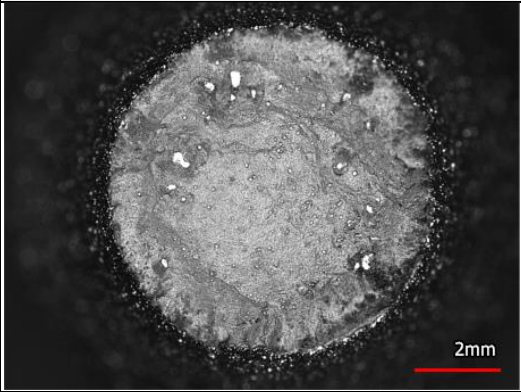
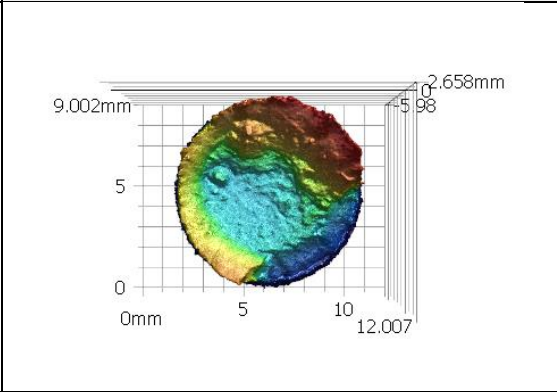
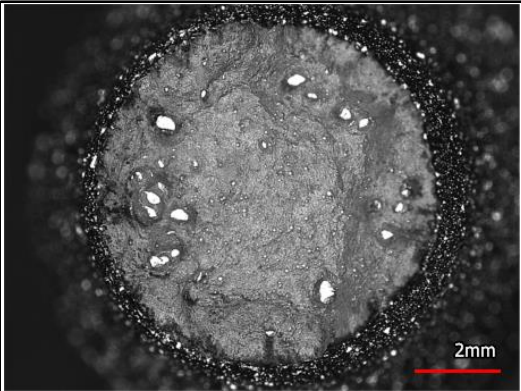
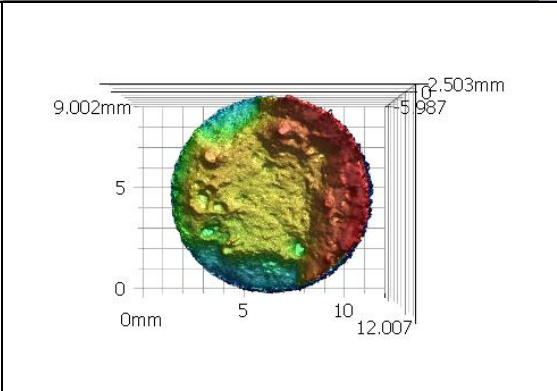
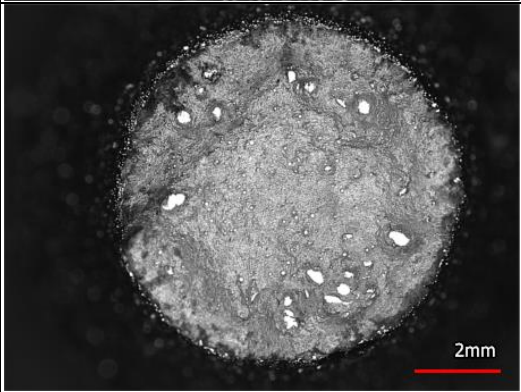
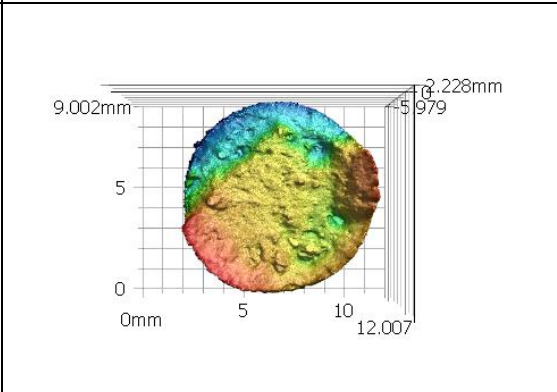


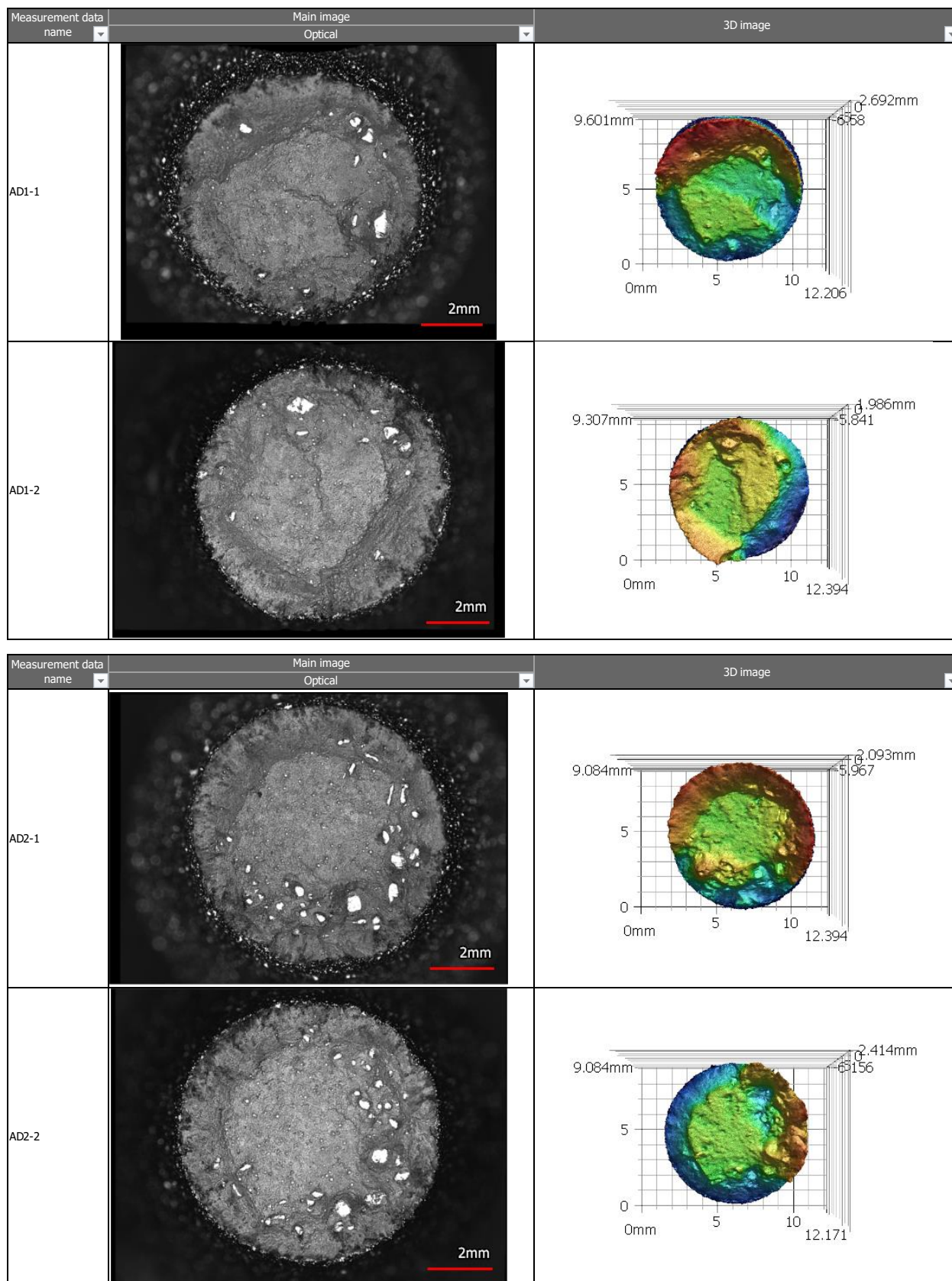
Measurement data name	Main image Optical	3D image
AG6-1		
AG6-2		
AG6-2-spheres		

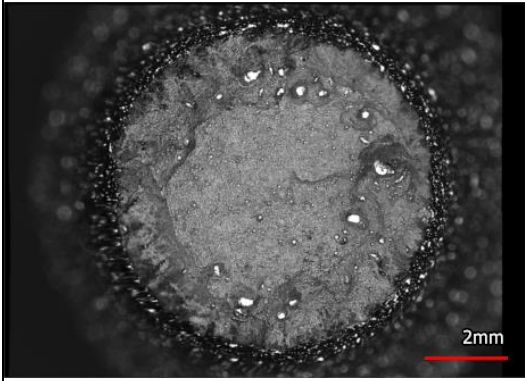
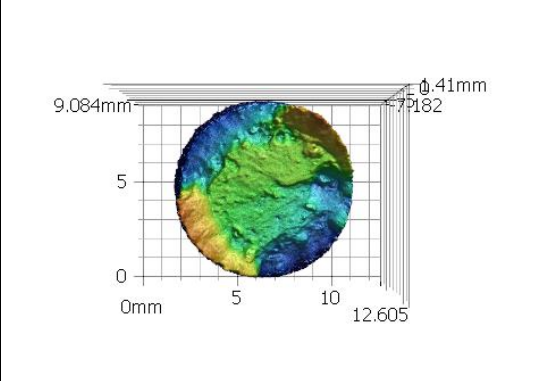
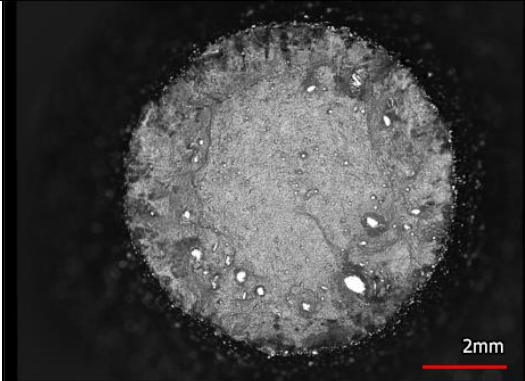
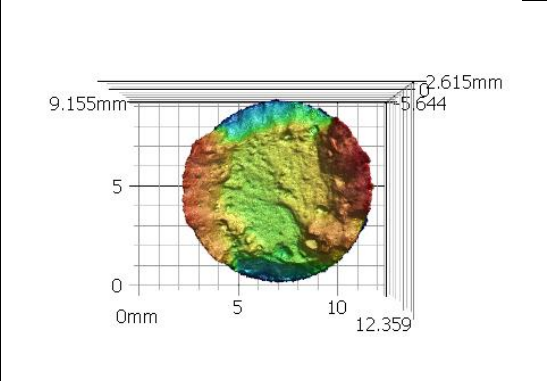
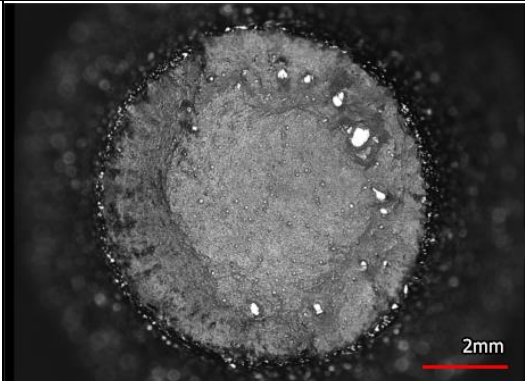
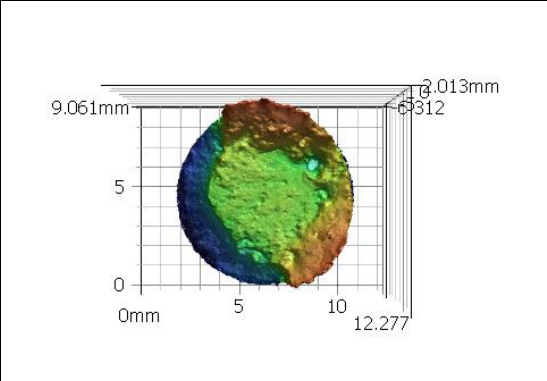
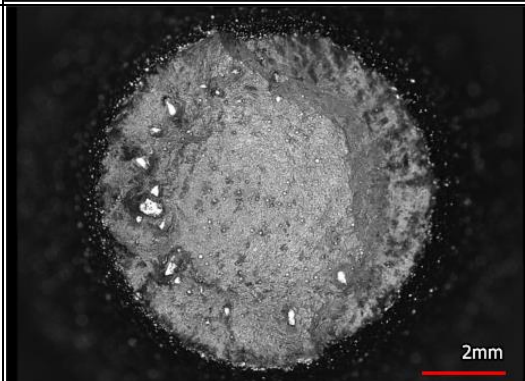
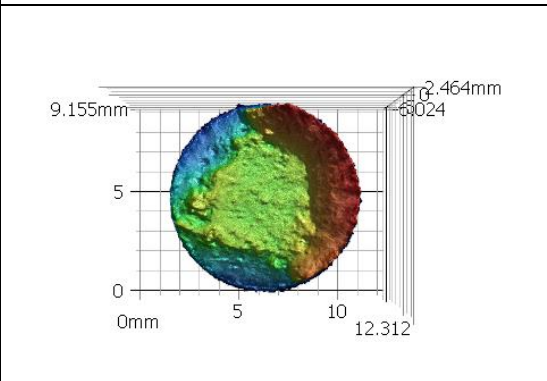
Measurement data name	Main image Optical	3D image
AP1-1		
AP1-2		
AP1-2-powder		
AP1-2-powder2		
AP1-2-powder3		

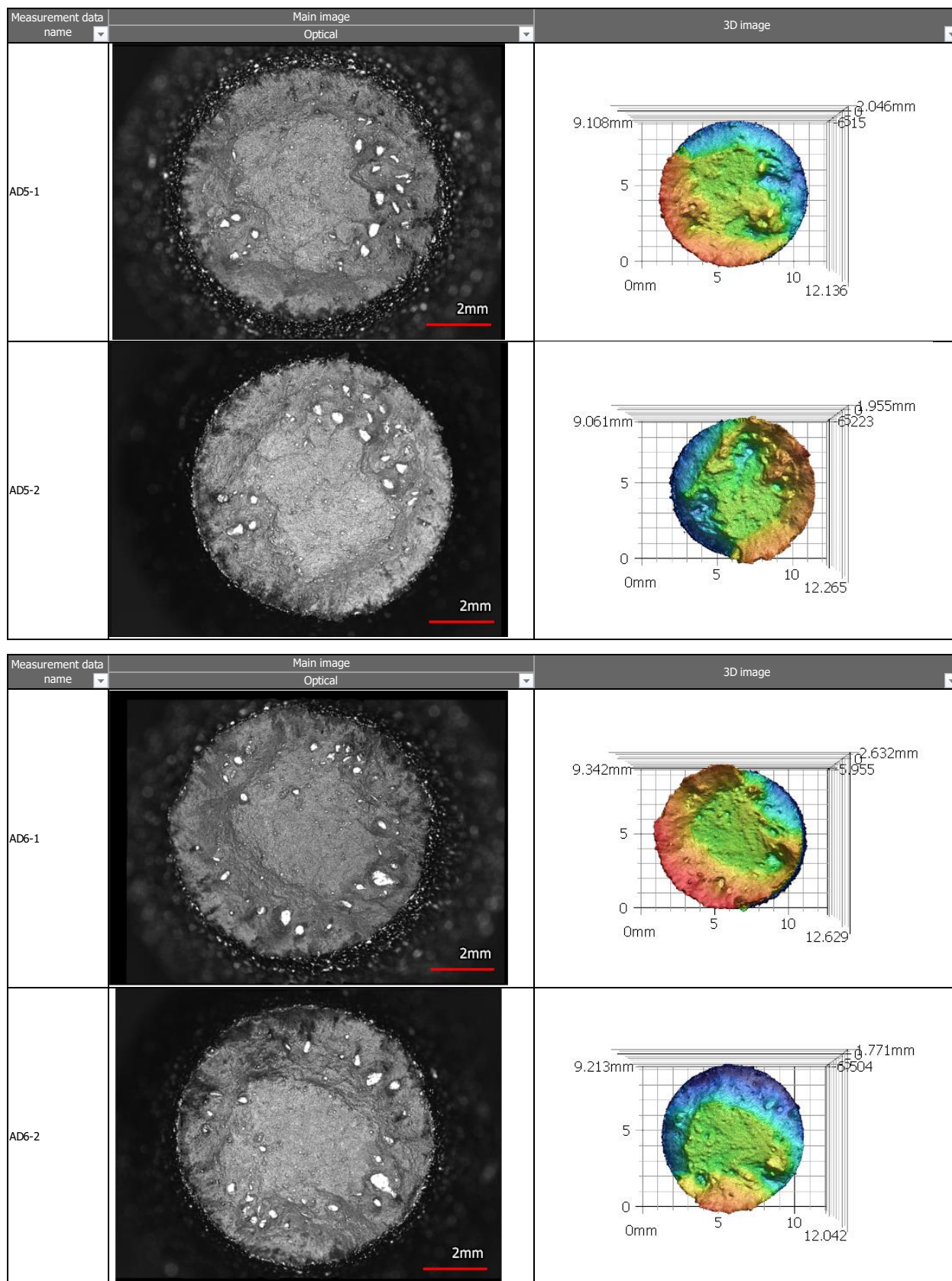
Measurement data name	Main image Optical	3D image
AP2-1		
AP2-2		
AP3-1		
AP3-2		

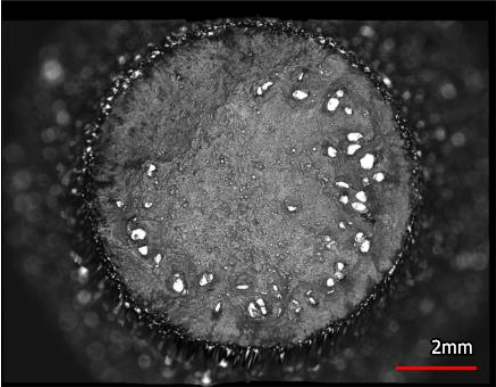
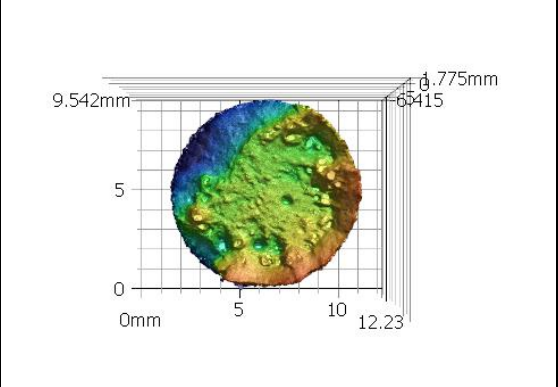
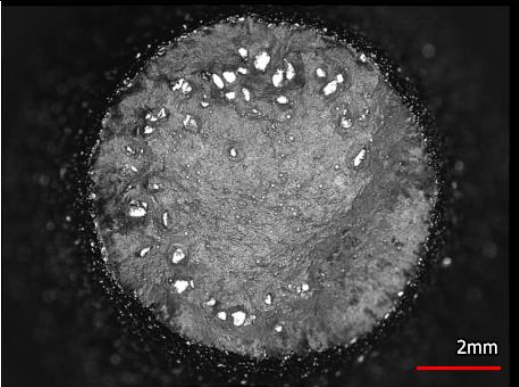
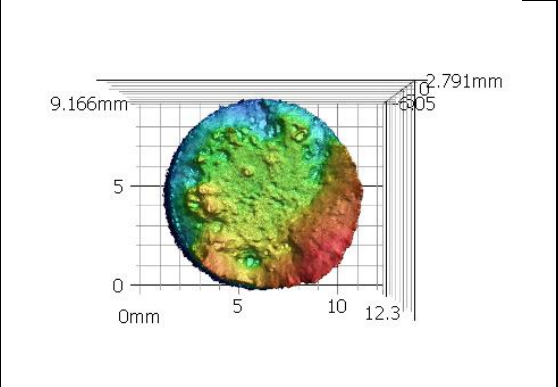
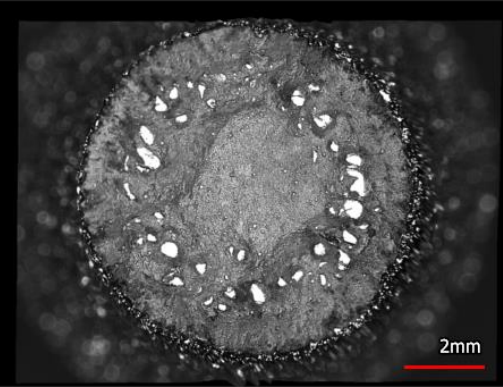
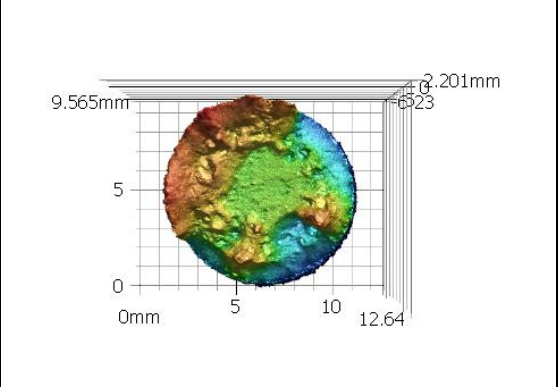
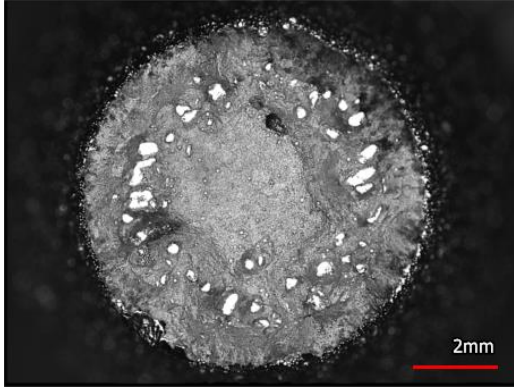
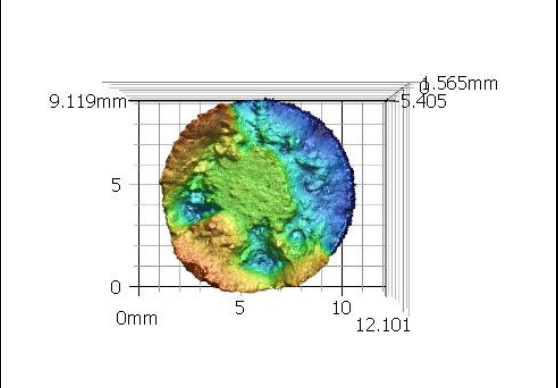
Measurement data name	Main image Optical	3D image
AP4-1		
AP4-2		
Measurement data name	Main image Optical	3D image
AP5-1		
AP5-2		

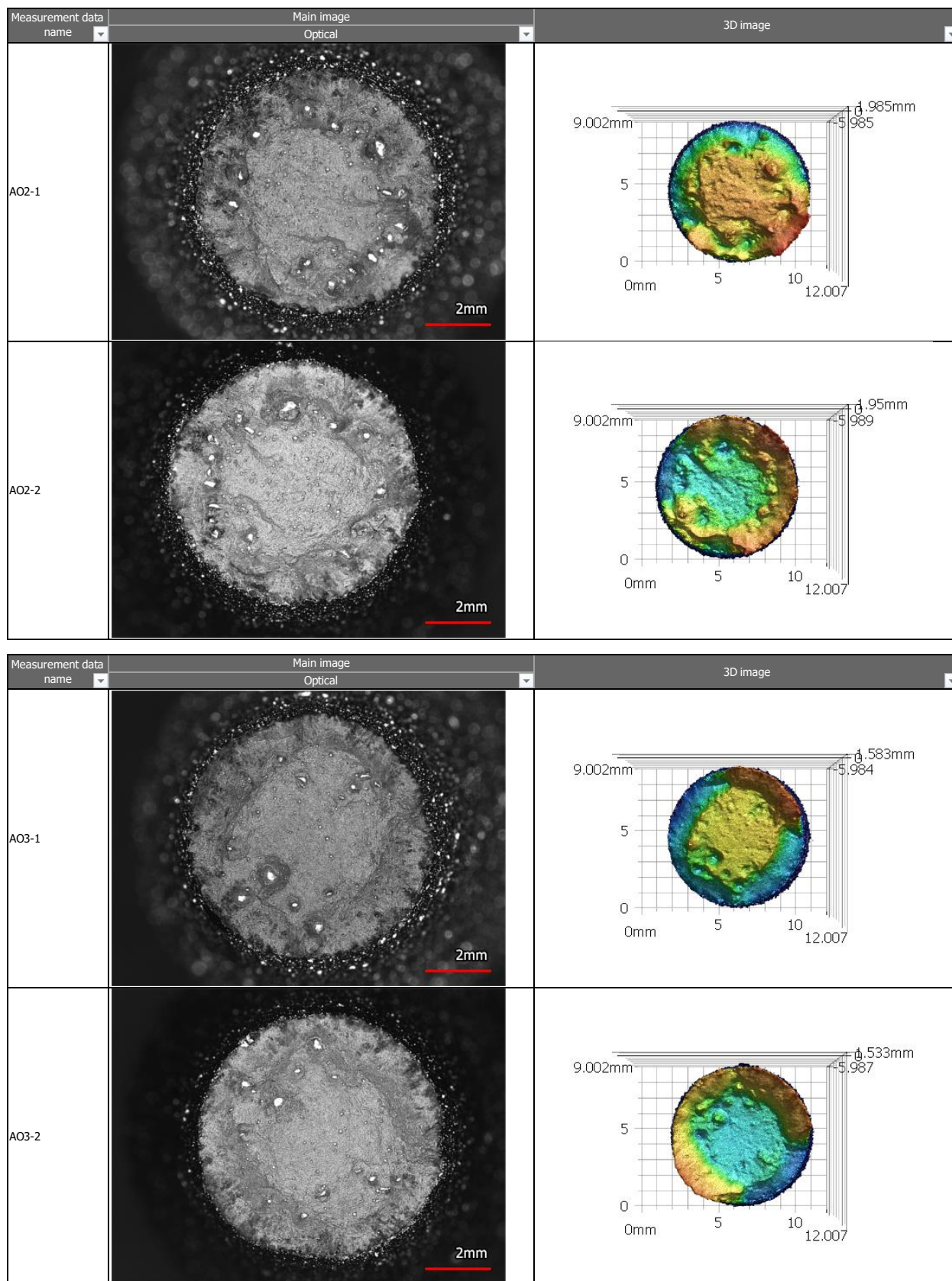
Measurement data name	Main image Optical	3D image
AP6-1		
AP6-2		
AP7-1		
AP7-2		

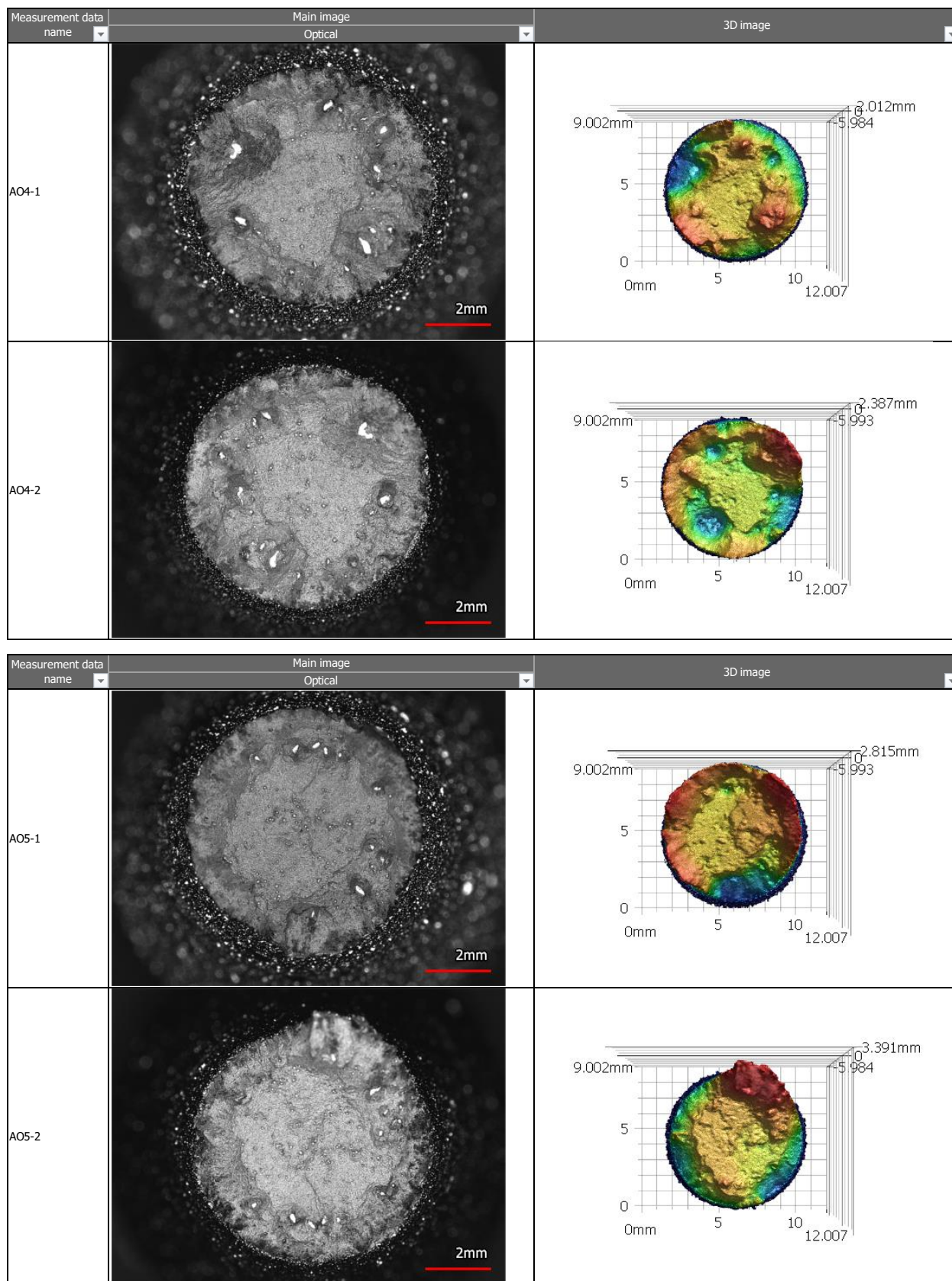


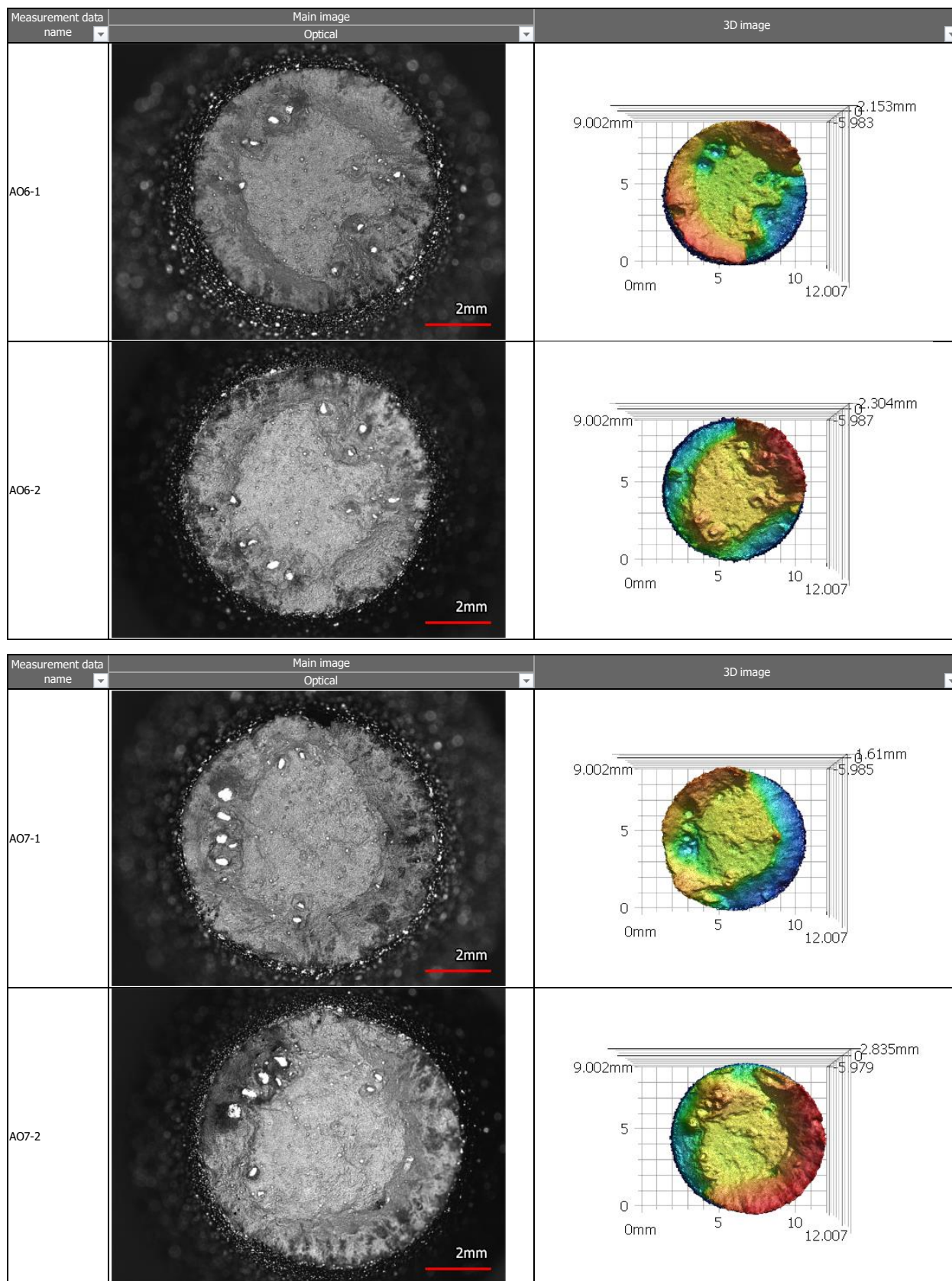
Measurement data name	Main image Optical	3D image
AD3-1		
AD3-2		
AD4-1		
AD4-2		

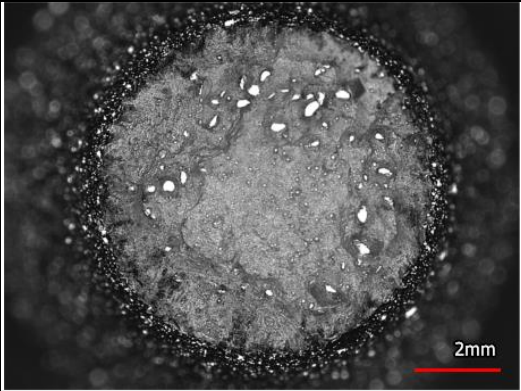
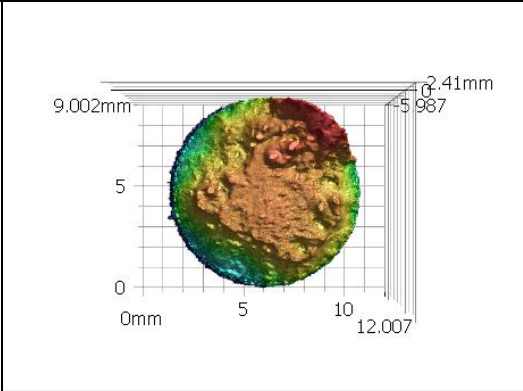
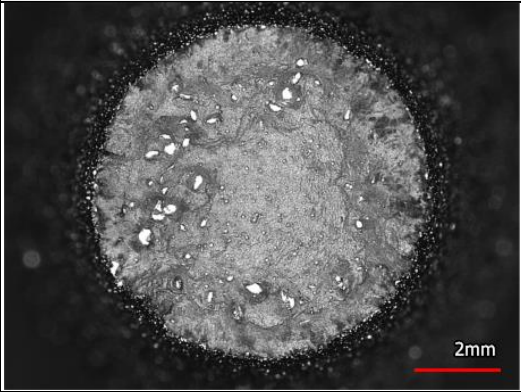
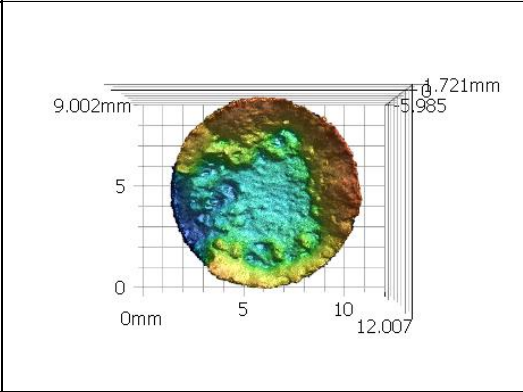
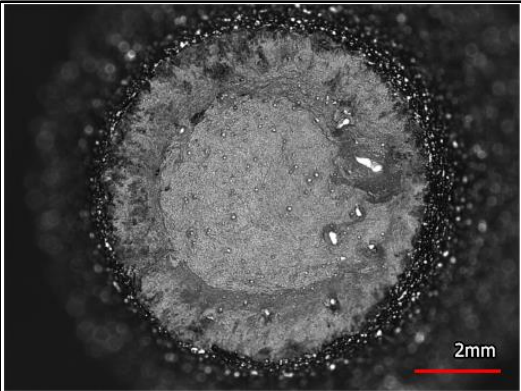
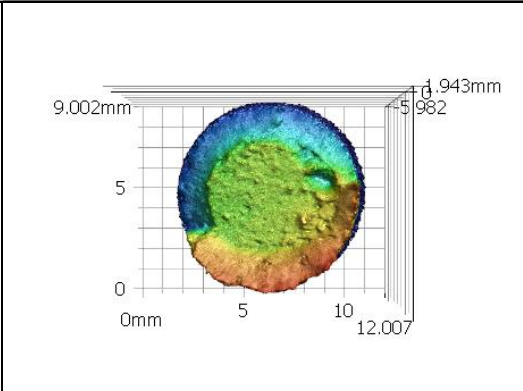
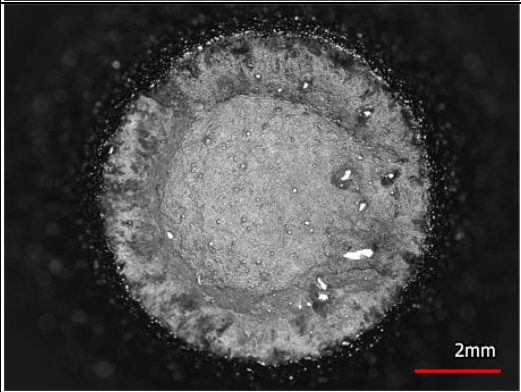
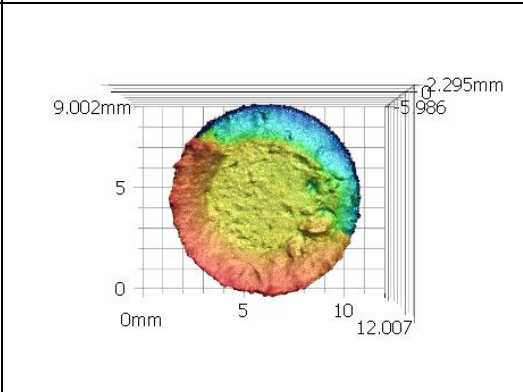


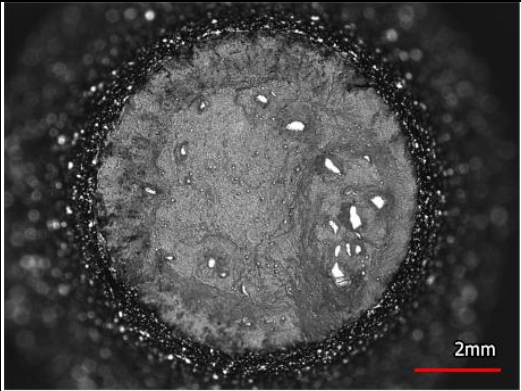
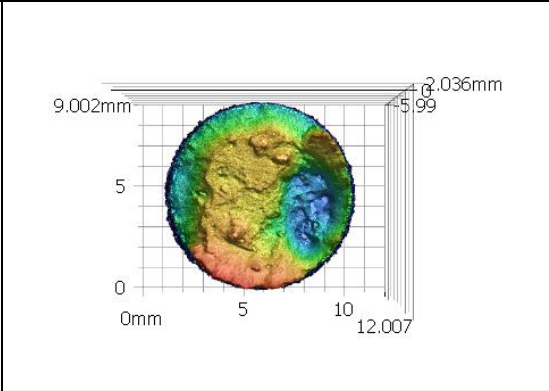
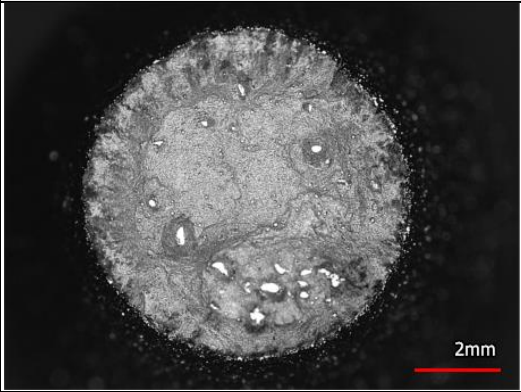
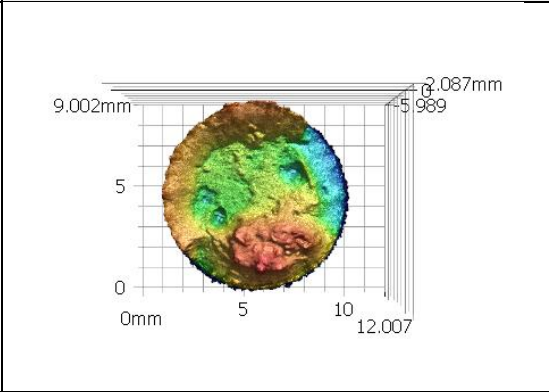
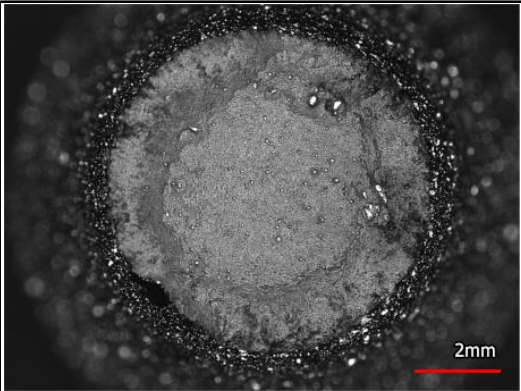
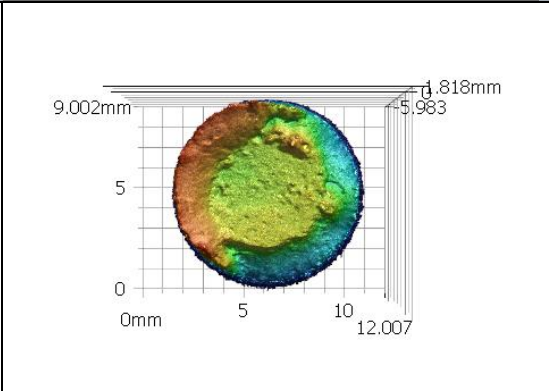
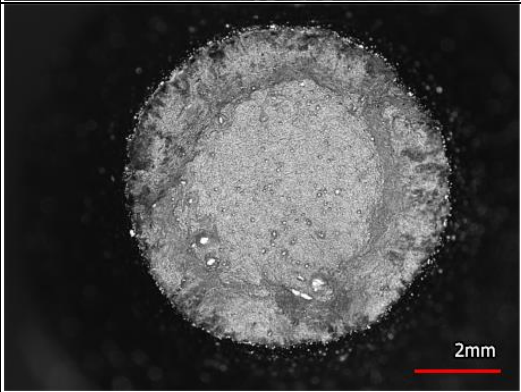
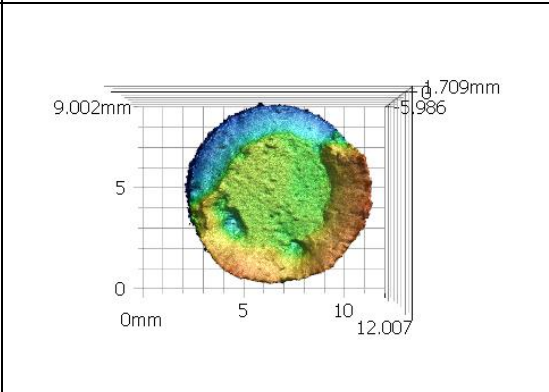
Measurement data name	Main image Optical	3D image
AD7-1		
AD7-2		
Measurement data name	Main image Optical	3D image
AO1-1		
AO1-2		

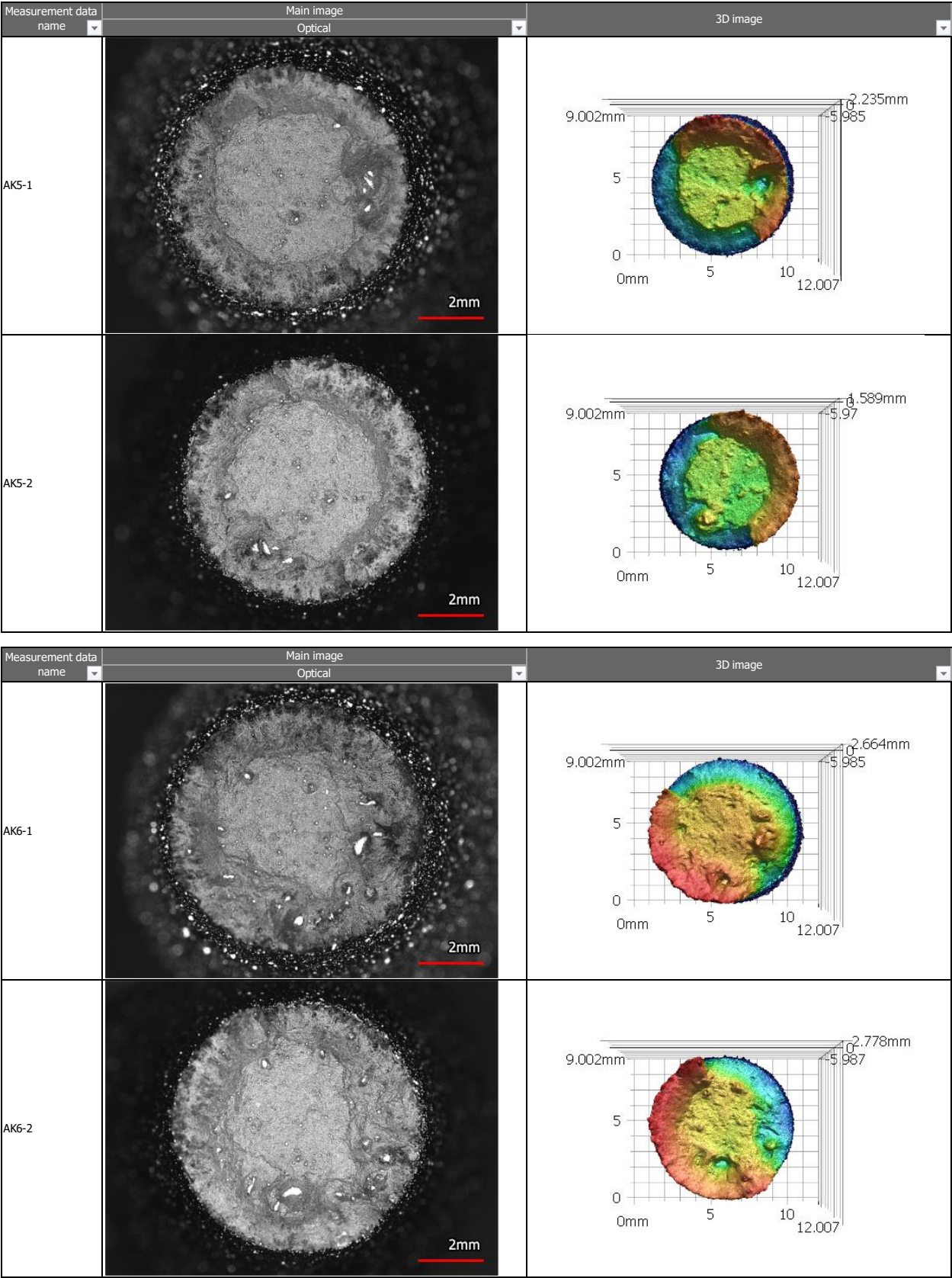


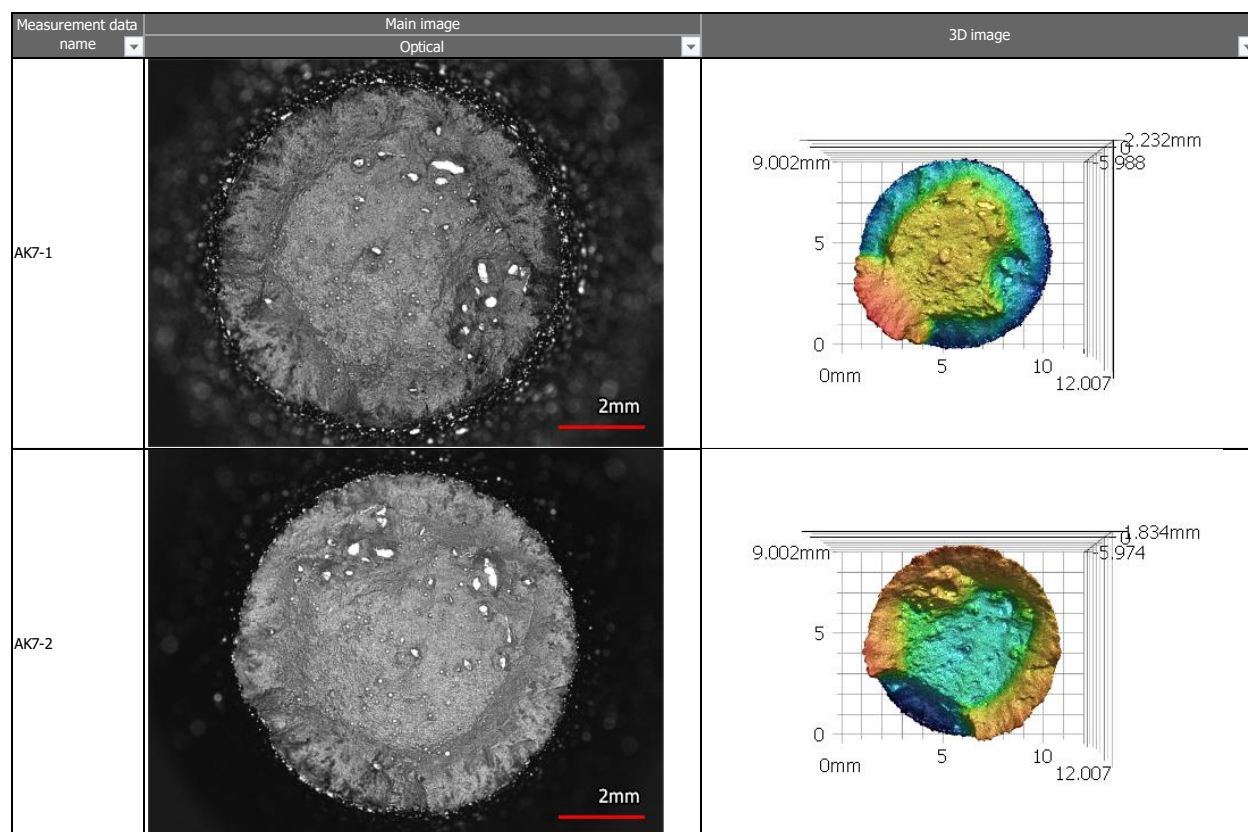




Measurement data name	Main image Optical	3D image
AK1-1		
AK1-2		
AK2-1		
AK2-2		

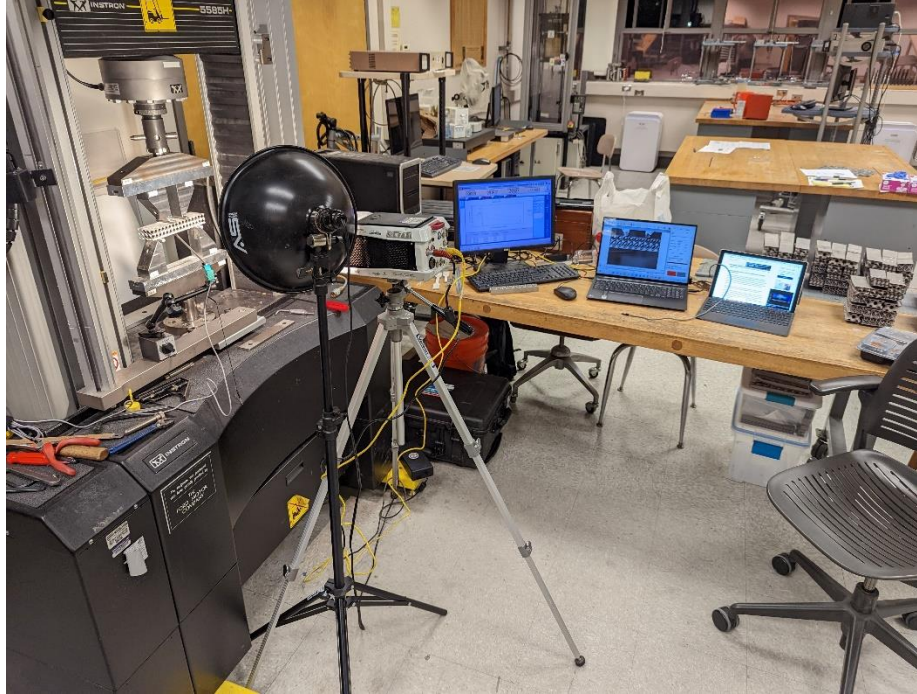
Measurement data name	Main image Optical	3D image
AK3-1		
AK3-2		
AK4-1		
AK4-2		





APPENDIX E

Beam flexure lab setup



Beam flexure property data

Grading	Build	Lattice Type	Specimen	Max Load	Modulus (Flexure)	Yield Stress (Flexure 0.2% Offset)	Yield Strain (Flexure 0.2% Offset)	Ultimate (Flexure) Strength	Peak Flexure Strain at max Load	Toughness
				N	MPa	MPa	mm/mm	MPa	mm/mm	MJ/mm ³
Uniform	1	Gyroid	U1_G	34109.63	10164.98	134.01	0.01552	138.46803	0.02164	6.20
Uniform	1	Primitive	U1_P	25945.18	11184.65	105.32	0.01049	105.32442	0.01049	3.58
Uniform	1	Diamond	U1_D	31838.09	9590.273	128.70	0.01647	129.24669	0.01759	6.32
Uniform	1	Octet	U1_O	20754.07	9877.418	84.25	0.01044	84.25115	0.01044	3.80
Uniform	1	Kelvin	U1_K	21770.53	4450.65	68.38	0.01777	88.37746	0.03746	4.70
Uniform	2	Gyroid	U2_G	35089.04	10474.68	139.02	0.01594	142.44392	0.02053	7.67
Uniform	2	Primitive	U2_P	25823.78	10728.83	104.83	0.01047	104.8316	0.01047	2.86
Uniform	2	Diamond	U2_D	39882.05	10418.54	157.11	0.01748	161.90114	0.0207	6.85
Uniform	2	Octet	U2_O	20106.19	9844.096	81.62	0.01048	81.62106	0.01048	6.67
Uniform	2	Kelvin	U2_K	23114.28	4445.292	73.43	0.01872	93.83239	0.03472	3.87
Uniform	3	Gyroid	U3_G	33555.28	9937.201	131.86	0.01623	136.21764	0.02057	5.14
Uniform	3	Primitive	U3_P	26607.58	11089.21	108.01	0.01082	108.01345	0.01082	4.62

Uniform	3	Diamond	U3_D	33198.66	9043.869	133.58	0.01786	134.76996	0.0205	6.50
Uniform	3	Octet	U3_O	27346.22	9270.166	111.01	0.01364	111.01196	0.01364	3.59
Uniform	3	Kelvin	U3_K	21382.63	4713.172	71.12	0.01779	86.80276	0.0415	4.12
Uniform	4	Gyroid	U4_G	34594.82	10148.28	136.90	0.01593	140.43764	0.02042	5.58
Uniform	4	Primitive	U4_P	26018.48	10341.15	105.62	0.01117	105.62201	0.01117	4.78
Uniform	4	Diamond	U4_D	33716.8	9357.101	133.55	0.017	136.87332	0.02066	6.71
Uniform	4	Octet	U4_O	21811.95	10096.67	88.55	0.01141	88.54559	0.01141	3.97
Uniform	4	Kelvin	U4_K	22286.42	4464.523	70.81	0.01835	90.47172	0.03521	4.46
Uniform	V	Gyroid	UV_G	33168.23	9432.603	132.58	0.01671	134.64642	0.01854	4.00
Uniform	V	Primitive	UV_P	22158.23	9506.455	89.95	0.01122	89.95131	0.01122	3.69
Uniform	V	Diamond	UV_D	33929.8	8933.299	136.47	0.01781	137.73801	0.01906	4.92
Uniform	V	Octet	UV_O	28517.87	9903.905	115.77	0.01418	115.76829	0.01418	4.68
Uniform	V	Kelvin	UV_K	20701.28	4672.474	76.26	0.01877	84.03683	0.03937	3.68
Uniform	A	Gyroid	UA_G	26149.05	8390.023	105.05	0.01526	106.15205	0.01759	6.05
Uniform	A	Primitive	UA_P	23285.83	9795.025	86.26	0.01008	94.52879	0.02525	4.76
Uniform	A	Diamond	UA_D	31064.74	8171.153	121.97	0.01756	126.10728	0.02253	7.26
Uniform	A	Octet	UA_O	22961.14	8815.623	93.21	0.01243	93.21073	0.01243	3.54
Uniform	A	Kelvin	UA_K	19386.33	4000.917	66.04	0.01956	78.69878	0.0315	4.42
Density	1	Gyroid	D1_G	36209.86	10055.83	139.94	0.01657	146.9939	0.02467	3.42
Density	1	Primitive	D1_P	26396.92	10998.48	107.16	0.01076	107.15826	0.01076	4.78
Density	1	Diamond	D1_D	38165.44	9864.032	148.45	0.01737	154.93259	0.02332	7.54
Density	1	Octet	D1_O	28095.21	10051.42	114.05	0.01347	114.05251	0.01347	4.60
Density	1	Kelvin	D1_K	28653.68	6176.142	92.16	0.01718	116.3196	0.03305	4.56
Density	2	Gyroid	D2_G	37282.8	10366.39	143.58	0.01654	151.3495	0.02474	4.29
Density	2	Primitive	D2_P	26690.69	11355.97	108.35	0.01095	108.35084	0.01095	4.67
Density	2	Diamond	D2_D	33323.95	9678.984	134.36	0.01659	135.27853	0.01807	5.95
Density	2	Octet	D2_O	26592.54	10267.52	107.95	0.01254	107.95238	0.01254	5.38
Density	2	Kelvin	D2_K	30909.96	5959.41	116.90	0.02192	125.47894	0.02534	4.01
Density	3	Gyroid	D3_G	36542.96	10432.05	143.49	0.01621	148.3461	0.02221	4.72
Density	3	Primitive	D3_P	26540.62	11184.68	107.74	0.02154	107.74163	0.02154	5.76
Density	3	Diamond	D3_D	38635.86	10043.26	149.62	0.01768	156.84222	0.02716	5.15
Density	3	Octet	D3_O	25339.54	10223.65	102.87	0.01254	102.86584	0.01254	5.43
Density	3	Kelvin	D3_K	30501.93	6405.164	92.94	0.01755	123.82258	0.03471	4.45
Density	4	Gyroid	D4_G	36886.57	10214.99	144.51	0.01676	149.74103	0.02442	3.93
Density	4	Primitive	D4_P	26964.12	11675.92	109.46	0.02181	109.46082	0.02181	5.49
Density	4	Diamond	D4_D	38741.6	10446.46	149.78	0.01704	157.27148	0.02452	7.18
Density	4	Octet	D4_O	28753.08	11382.93	116.72	0.01361	116.72311	0.01361	5.69
Density	4	Kelvin	D4_K	30774.41	5852.626	99.02	0.01951	124.9287	0.03684	6.15
Density	V	Gyroid	DV_G	34827.81	9397.023	138.15	0.01777	141.38348	0.0178	3.71
Density	V	Primitive	DV_P	23293.3	9297.131	94.56	0.02501	94.55912	0.02501	3.10
Density	V	Diamond	DV_D	39356.29	9491.18	148.56	0.0186	159.76682	0.02968	4.85
Density	V	Octet	DV_O	31039.29	9221.646	126.00	0.0159	126.00397	0.0159	2.01
Density	V	Kelvin	DV_K	29720.95	5769.05	120.55	0.02375	120.65217	0.02395	3.19
Density	A	Gyroid	DA_G	33042.04	9629.504	129.93	0.01679	134.13412	0.03647	5.90
Density	A	Primitive	DA_P	26149.81	10265.24	96.52	0.01189	106.15512	0.02658	5.42

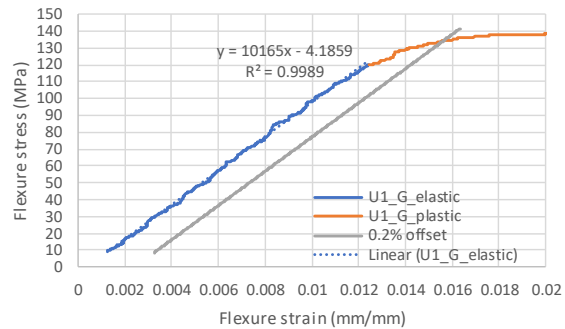
Density	A	Diamond	DA_D	38086.82	9121.182	139.16	0.017	154.61343	0.02989	7.98
Density	A	Octet	DA_O	32683.03	10156.14	128.32	0.01525	132.67674	0.01819	3.28
Density	A	Kelvin	DA_K	29446	5977.309	103.60	0.01982	119.536	0.05087	6.90
Stress	1	Gyroid	S1_G	40525.63	11789.74	152.82	0.01575	164.51375	0.02417	3.07
Stress	1	Primitive	S1_P	28601.1	12124.97	116.11	0.02137	116.10614	0.02137	5.89
Stress	1	Diamond	S1_D	48700.05	12015.72	178.44	0.01758	197.6978	0.02914	4.76
Stress	1	Octet	S1_O	35106.95	11920.75	142.52	0.01456	142.51665	0.01456	6.26
Stress	1	Kelvin	S1_K	28714.33	6668.79	114.67	0.0196	116.5658	0.02613	3.27
Stress	3	Gyroid	S3_G	41899.94	11890.19	160.53	0.01629	170.09276	0.02386	2.76
Stress	3	Primitive	S3_P	28914.55	12055.14	117.38	0.02169	117.37861	0.02169	5.83
Stress	3	Diamond	S3_D	50102.43	11970.3	185.22	0.01846	203.39078	0.02943	4.91
Stress	3	Octet	S3_O	35529.07	12183.92	143.89	0.0148	144.23026	0.01523	6.42
Stress	3	Kelvin	S3_K	30096.73	6453.144	116.39	0.02069	122.17765	0.02571	3.19
Stress	4	Gyroid	S4_G	42746.19	11819.87	162.97	0.01644	173.52814	0.02592	3.30
Stress	4	Primitive	S4_P	29788.96	12718.26	120.93	0.02308	120.92825	0.02308	5.94
Stress	4	Diamond	S4_D	50330.31	12019.06	186.44	0.01791	204.31586	0.03314	5.19
Stress	4	Octet	S4_O	37575.33	12363.79	151.00	0.01525	152.53702	0.01684	6.92
Stress	4	Kelvin	S4_K	29932.39	6770.676	115.40	0.01974	121.51053	0.02461	3.84
Stress	V	Gyroid	SV_G	38963.19	10636.66	157.39	0.01745	158.17105	0.01761	1.48
Stress	V	Primitive	SV_P	25048.61	10995.02	101.68	0.01329	101.68483	0.01329	3.66
Stress	V	Diamond	SV_D	45406.91	10995.08	183.02	0.01939	184.32933	0.01966	1.96
Stress	V	Octet	SV_O	38955.29	11549.27	153.51	0.01628	158.13898	0.0202	4.00
Stress	V	Kelvin	SV_K	28957.81	6576.432	115.62	0.02038	117.55422	0.02198	2.67
Stress	A	Gyroid	SA_G	39827.55	10776.54	153.08	0.01724	161.6799	0.02484	2.80
Stress	A	Primitive	SA_P	28624.66	10953.14	106.53	0.01106	116.20179	0.02768	4.24
Stress	A	Diamond	SA_D	46917.95	10710.6	168.21	0.01895	190.46339	0.0396	6.23
Stress	A	Octet	SA_O	39065.17	11279.03	150.11	0.01603	158.58504	0.02136	7.70
Stress	A	Kelvin	SA_K	28383.46	6326.617	114.64	0.02116	115.22263	0.02868	3.80

Beam flexure property calculations for averages and standard deviation

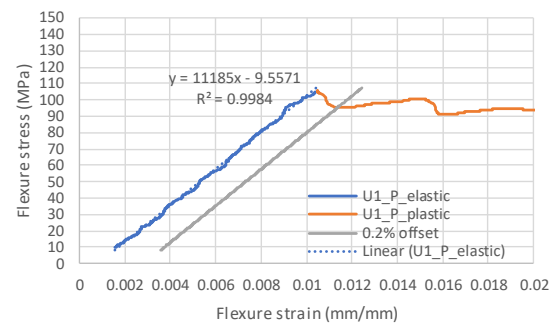
				Max Load	Modulus (Flexure)	Yield Stress (Flexure 0.2% Offset)	Yield Strain (Flexure 0.2% Offset)	Ultimate (Flexure) Strength	Peak Flexure Strain at max Load	Toughness
				N	MPa	MPa	mm/mm	MPa	mm/mm	MJ/mm ³
Uniform	Gyroid	U_G	Average	34337.19	10181.29	135.4456575	0.015905	139.39181	0.02079	6.14736954
			Std Dev	568.949	191.7112	2.729291857	0.000252834	2.3096461	0.000494	0.95798483
Density	Gyroid	D_G	Average	36730.55	10267.31	142.8794375	0.01652	149.10763	0.02401	4.08865077
			Std Dev	398.6334	145.2734	1.741036247	0.000197864	1.6182521	0.001046	0.47550449
Stress	Gyroid	S_G	Average	41723.92	11833.27	158.7729367	0.01616	169.37822	0.02465	3.04330755
			Std Dev	915.0468	42.08639	4.325638003	0.000296311	3.714632	0.000907	0.22027591
Uniform	Primitive	U_P	Average	26098.75	10835.96	105.94787	0.0107375	105.94787	0.010738	3.95797653
			Std Dev	301.8886	332.427	1.225515615	0.000285777	1.2255156	0.000286	0.78400237
Density	Primitive	D_P	Average	26648.09	11303.76	108.1778875	0.016265	108.17789	0.016265	5.17518802

			Std Dev	209.9566	249.3048	0.85231878	0.005411259	0.8523188	0.005411	0.46263605
Stress	Primitive	S_P	Average	29101.54	12299.46	118.1376667	0.022046667	118.13767	0.022047	5.88677673
			Std Dev	502.642	297.5095	2.040475677	0.000742264	2.0404757	0.000742	0.045251
Uniform	Diamond	U_D	Average	34658.9	9602.445	138.23546	0.0172025	140.69778	0.019863	6.59311371
			Std Dev	3092.656	509.498	11.07896136	0.000521266	12.554635	0.001314	0.20290679
Density	Diamond	D_D	Average	37216.71	10008.19	145.5497925	0.01717	151.08121	0.023268	6.45380435
			Std Dev	2257.929	283.9325	6.483430068	0.000404166	9.1660583	0.003307	0.95393335
Stress	Diamond	S_D	Average	49710.93	12001.69	183.3669167	0.017983333	201.80148	0.03057	4.95286271
			Std Dev	720.8297	22.24136	3.515735272	0.000362981	2.9262132	0.001821	0.17801362
Uniform	Octet	U_O	Average	22504.61	9772.088	91.35744	0.0114925	91.35744	0.011493	4.50647077
			Std Dev	2860.847	305.5979	11.61360312	0.00129918	11.613603	0.001299	1.25823212
Density	Octet	D_O	Average	27195.09	10481.38	110.39846	0.01304	110.39846	0.01304	5.27655311
			Std Dev	1326.998	526.7364	5.386950679	0.000502444	5.3869507	0.000502	0.41014546
Stress	Octet	S_O	Average	36070.45	12156.15	145.8011567	0.01487	146.42798	0.015543	6.53326683
			Std Dev	1077.972	181.9331	3.7180305	0.000286007	4.3760273	0.000957	0.27871274
Uniform	Kelvin	U_K	Average	22138.47	4518.409	70.9338475	0.0181575	89.871083	0.037223	4.28987745
			Std Dev	648.2216	112.6652	1.791103111	0.000399586	2.6314556	0.002677	0.31706276
Density	Kelvin	D_K	Average	30210	6098.335	100.25566	0.01904	122.63746	0.032485	4.7906003
			Std Dev	910.4735	212.056	9.968821569	0.001883813	3.6960673	0.004338	0.80998442
Stress	Kelvin	S_K	Average	29581.15	6630.87	115.4901233	0.02001	120.08466	0.025483	3.43447184
			Std Dev	616.5953	132.3758	0.705441724	0.000484218	2.5030707	0.000641	0.28527158

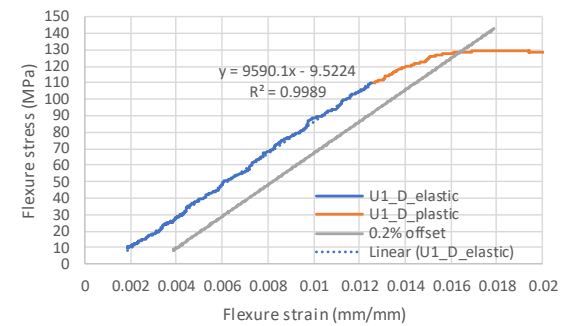
End date	10/15/2022									
Specimen label	U1_G									
Support span	142 mm									
Loading span	47.3 mm									
Thickness (Sandwich)	27.0256 mm									
Width	48 mm									
Maximum Load	34109.63281 N									
Flexure extension at Maximum Load	-3.44167 mm									
Flexure strain (Extension) at Maximum Load	-0.02164 mm/mm									
Flexure stress at Maximum Load	138.46803 MPa									
Rate 1	1									
(Cube) Core Thickness	25.4 mm									
Facesheet Thickness	0.8128 mm									
Modulus	10164.9777 MPa									
Linear	409	1745								
0.2% offset Yield strain	0.01552 mm/mm									
0.2% offset Yield stress	134.01 MPa									
Toughness	6.20 MJ/mm ³									



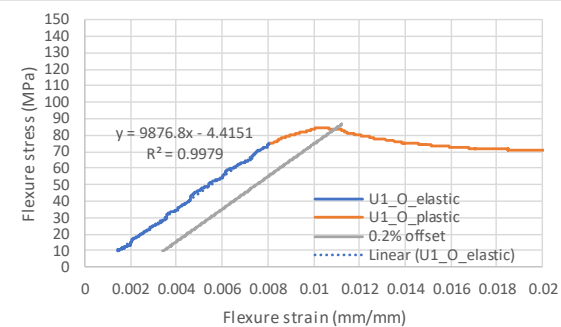
End date	10/15/2022
Specimen label	U1_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	25945.17578 N
Flexure extension at Maximum Load	-1.66865 mm
Flexure strain (Extension) at Maximum Load	-0.01049 mm/mm
Flexure stress at Maximum Load	105.32442 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11184.64961 MPa
Linear	555 1550
0.2% offset Yield strain	0.01049 mm/mm
0.2% offset Yield stress	105.32 MPa
Toughness	3.58 MJ/mm ³



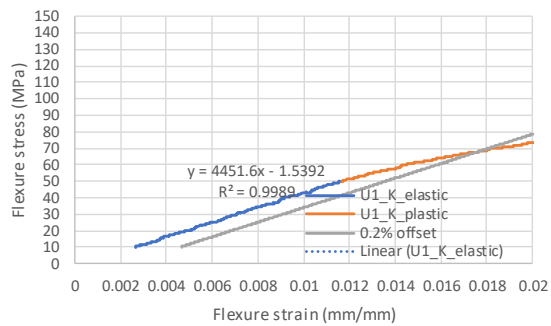
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Specimen label	U1_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	31838.08594 N
Flexure extension at Maximum Load	-2.79687 mm
Flexure strain (Extension) at Maximum Load	-0.01759 mm/mm
Flexure stress at Maximum Load	129.24669 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9590.272726 MPa
Linear	437 1711
0.2% offset Yield strain	0.01647 mm/mm
0.2% offset Yield stress	128.70 MPa
Toughness	6.32 MJ/mm ³



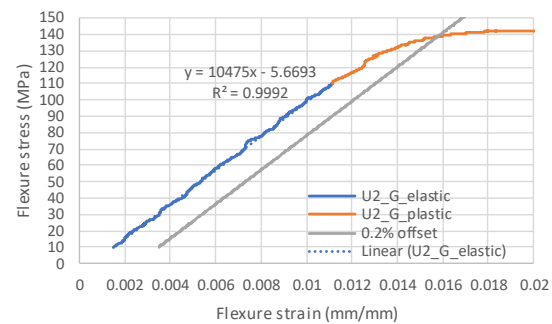
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Specimen label	U1_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	20754.07422 N
Flexure extension at Maximum Load	-1.65955 mm
Flexure strain (Extension) at Maximum Load	-0.01044 mm/mm
Flexure stress at Maximum Load	84.25115 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9877.418224 MPa
Linear	401 1341
0.2% offset Yield strain	0.01044 mm/mm
0.2% offset Yield stress	84.25 MPa
Toughness	3.80 MJ/mm ³



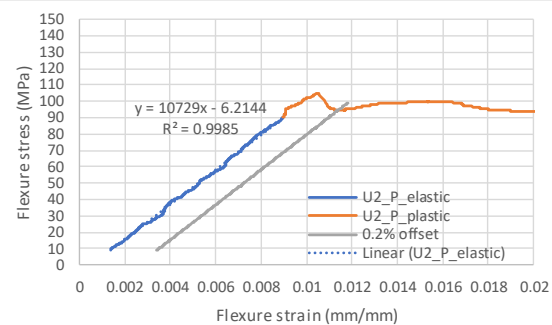
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Specimen label	U1_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	21770.53125 N
Flexure extension at Maximum Load	-5.95777 mm
Flexure strain (Extension) at Maximum Load	-0.03746 mm/mm
Flexure stress at Maximum Load	88.37746 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	4450.649595 MPa
Linear	462 1479
0.2% offset Yield strain	0.01777 mm/mm
0.2% offset Yield stress	68.38 MPa
Toughness	4.70 MJ/mm ³



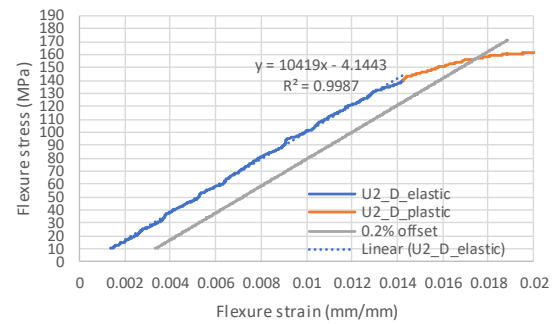
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Specimen label	U2_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	35089.03906 N
Flexure extension at Maximum Load	-3.26408 mm
Flexure strain (Extension) at Maximum Load	-0.02053 mm/mm
Flexure stress at Maximum Load	142.44392 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10474.68023 Pa
Linear	239 1379
0.2% offset Yield strain	0.01594 mm/mm
0.2% offset Yield stress	139.01532 MPa
Toughness	7.67 MJ/mm ³



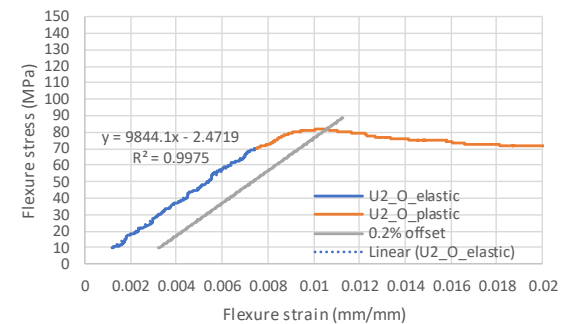
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Specimen label	U2_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	25823.77539 N
Flexure extension at Maximum Load	-1.66498 mm
Flexure strain (Extension) at Maximum Load	-0.01047 mm/mm
Flexure stress at Maximum Load	104.8316 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10728.82723 Pa
Linear	540 1399
0.2% offset Yield strain	0.01047 mm/mm
0.2% offset Yield stress	104.8316 MPa
Toughness	2.86 MJ/mm ³



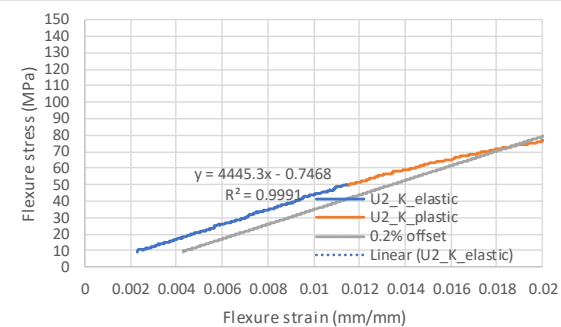
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Specimen label	D2_D was U2_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	39882.04688 N
Flexure extension at Maximum Load	-3.29154 mm
Flexure strain (Extension) at Maximum Load	-0.0207 mm/mm
Flexure stress at Maximum Load	161.90114 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10418.53576 Pa
Linear	295 1835
0.2% offset Yield strain	0.01748 mm/mm
0.2% offset Yield stress	157.11482 MPa
Toughness	6.85 MJ/mm ³



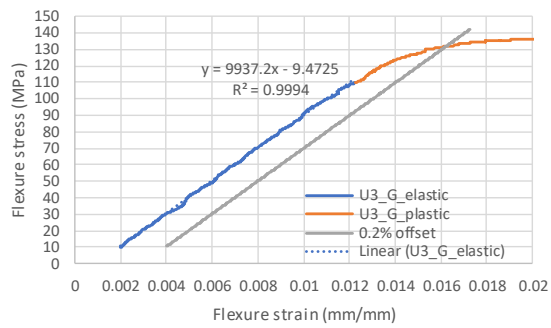
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Specimen label	U2_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	20106.18945 N
Flexure extension at Maximum Load	-1.66656 mm
Flexure strain (Extension) at Maximum Load	-0.01048 mm/mm
Flexure stress at Maximum Load	81.62106 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9844.096 Pa
Linear	311 1193
0.2% offset Yield strain	0.01048 mm/mm
0.2% offset Yield stress	81.62106 MPa
Toughness	6.67 MJ/mm ³



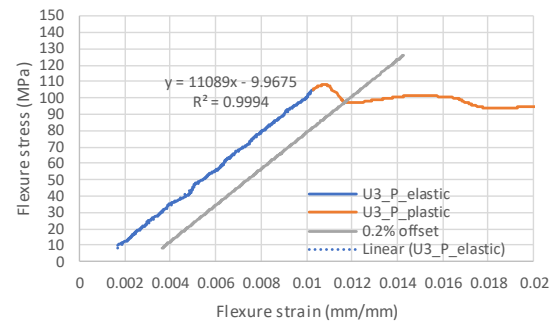
End date	10/14/2022
Specimen label	U2_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	23114.27734 N
Flexure extension at Maximum Load	-5.52188 mm
Flexure strain (Extension) at Maximum Load	-0.03472 mm/mm
Flexure stress at Maximum Load	93.83239 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	4445.291812 Pa
Linear	708 1696
0.2% offset Yield strain	0.01872 mm/mm
0.2% offset Yield stress	73.43102 MPa
Toughness	3.87 MJ/mm ³



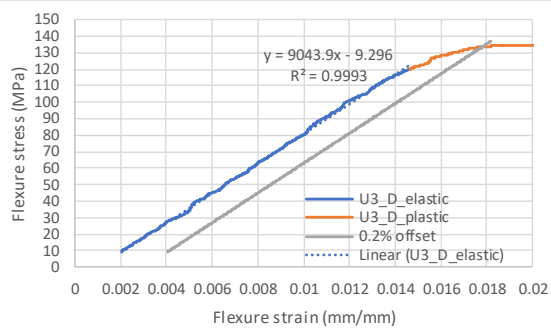
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Specimen label	U3_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	33555.28125 N
Flexure extension at Maximum Load	-3.2704 mm
Flexure strain (Extension) at Maximum Load	-0.02057 mm/mm
Flexure stress at Maximum Load	136.21764 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9937.201289 Pa
Linear	375 1590
0.2% offset Yield strain	0.01623 mm/mm
0.2% offset Yield stress	131.85674 MPa
Toughness	5.14 MJ/mm ³



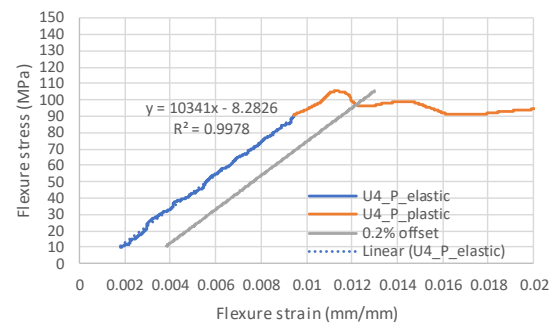
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Specimen label	U3_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26607.58008 N
Flexure extension at Maximum Load	-1.72076 mm
Flexure strain (Extension) at Maximum Load	-0.01082 mm/mm
Flexure stress at Maximum Load	108.01345 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11089.2093 Pa
Linear	363 1385
0.2% offset Yield strain	0.01082 mm/mm
0.2% offset Yield stress	108.01345 MPa
Toughness	4.62 MJ/mm ³



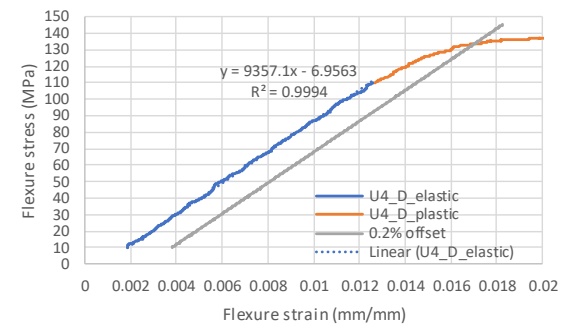
End date	10/15/2022
Specimen label	U3_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	33198.66406 N
Flexure extension at Maximum Load	-3.25948 mm
Flexure strain (Extension) at Maximum Load	-0.0205 mm/mm
Flexure stress at Maximum Load	134.76996 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9043.869339 Pa
Linear	491 1969
0.2% offset Yield strain	0.01786 mm/mm
0.2% offset Yield stress	133.57634 MPa
Toughness	6.50 MJ/mm ³



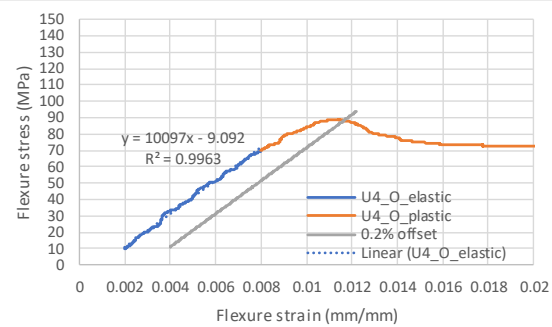
End date	10/16/2022
Specimen label	U4_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26018.48242 N
Flexure extension at Maximum Load	-1.77675 mm
Flexure strain (Extension) at Maximum Load	-0.01117 mm/mm
Flexure stress at Maximum Load	105.62201 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10341.15327 MPa
Linear	467 1351
0.2% offset Yield strain	0.01117 mm/mm
0.2% offset Yield stress	105.62 MPa
Toughness	4.78 MJ/mm ³



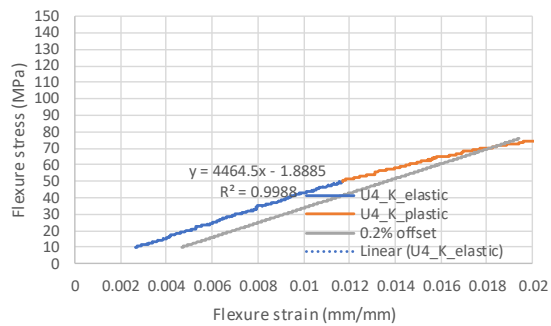
End date	10/15/2022
Specimen label	U4_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	33716.80078 N
Flexure extension at Maximum Load	-3.28612 mm
Flexure strain (Extension) at Maximum Load	-0.02066 mm/mm
Flexure stress at Maximum Load	136.87332 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9357.101078 MPa
Linear	309 1656
0.2% offset Yield strain	0.017 mm/mm
0.2% offset Yield stress	133.55 MPa
Toughness	6.71 MJ/mm ³



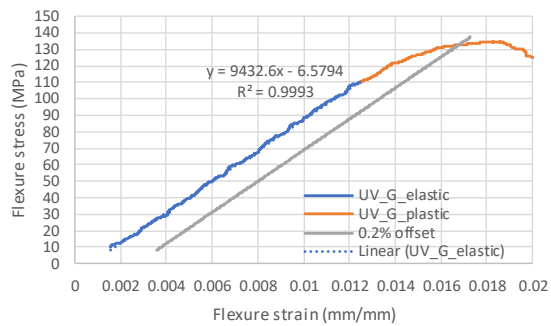
End date	10/16/2022
Specimen label	U4_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	21811.94727 N
Flexure extension at Maximum Load	-1.8148 mm
Flexure strain (Extension) at Maximum Load	-0.01141 mm/mm
Flexure stress at Maximum Load	88.54559 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10096.67165 MPa
Linear	365 1200
0.2% offset Yield strain	0.01141 mm/mm
0.2% offset Yield stress	88.55 MPa
Toughness	3.97 MJ/mm ³



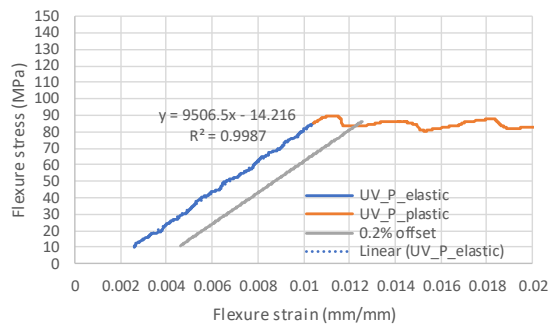
End date	10/15/2022
Specimen label	U4_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	22286.42383 N
Flexure extension at Maximum Load	-5.5992 mm
Flexure strain (Extension) at Maximum Load	-0.03521 mm/mm
Flexure stress at Maximum Load	90.47172 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	4464.522502 MPa
Linear	468 1514
0.2% offset Yield strain	0.01835 mm/mm
0.2% offset Yield stress	70.81 MPa
Toughness	4.46 MJ/mm ³



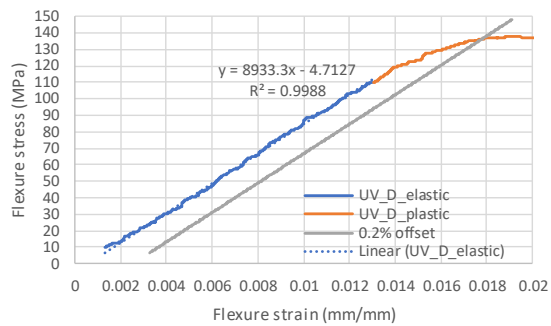
End date	10/16/2022
Specimen label	UV_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	33168.23438 N
Flexure extension at Maximum Load	-2.9489 mm
Flexure strain (Extension) at Maximum Load	-0.01854 mm/mm
Flexure stress at Maximum Load	134.64642 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9432.603325 MPa
Linear	276 1533
0.2% offset Yield strain	0.01671 mm/mm
0.2% offset Yield stress	132.58 MPa
Toughness	4.00 MJ/mm ³



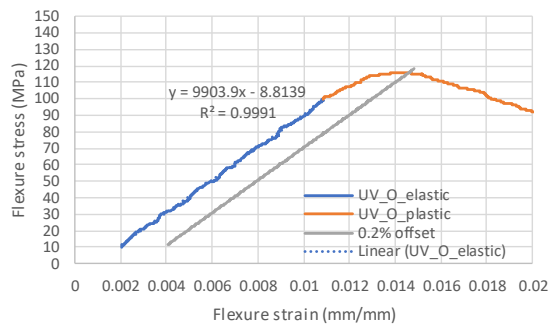
End date	10/16/2022
Specimen label	UV_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	22158.22852 N
Flexure extension at Maximum Load	-1.78404 mm
Flexure strain (Extension) at Maximum Load	-0.01122 mm/mm
Flexure stress at Maximum Load	89.95131 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9506.455015 MPa
Linear	501 1383
0.2% offset Yield strain	0.01122 mm/mm
0.2% offset Yield stress	89.95 MPa
Toughness	3.69 MJ/mm ³



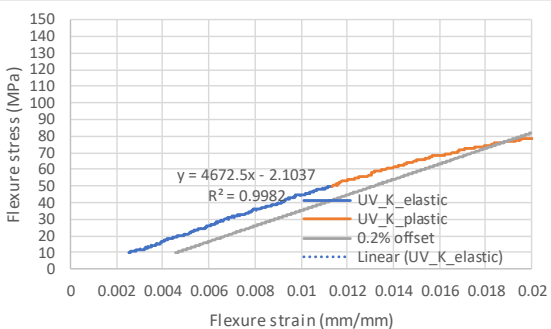
End date	10/16/2022
Specimen label	UV_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	33929.80078 N
Flexure extension at Maximum Load	-3.03106 mm
Flexure strain (Extension) at Maximum Load	-0.01906 mm/mm
Flexure stress at Maximum Load	137.73801 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	8933.299222 MPa
Linear	314 1628
0.2% offset Yield strain	0.01781 mm/mm
0.2% offset Yield stress	136.47 MPa
Toughness	4.92 MJ/mm ³



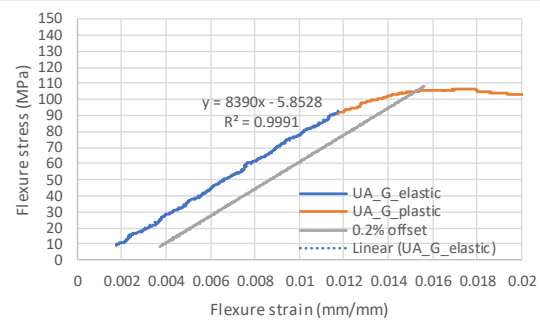
End date	10/16/2022
Specimen label	UV_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28517.87305 N
Flexure extension at Maximum Load	-2.25546 mm
Flexure strain (Extension) at Maximum Load	-0.01418 mm/mm
Flexure stress at Maximum Load	115.76829 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9903.904807 MPa
Linear	374 1450
0.2% offset Yield strain	0.01418 mm/mm
0.2% offset Yield stress	115.77 MPa
Toughness	4.68 MJ/mm ³



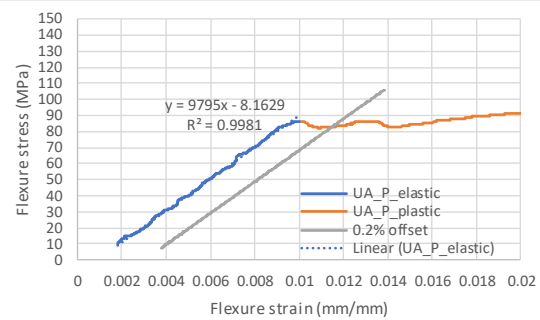
End date	10/16/2022
Specimen label	UV_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	20701.2793 N
Flexure extension at Maximum Load	-6.26043 mm
Flexure strain (Extension) at Maximum Load	-0.03937 mm/mm
Flexure stress at Maximum Load	84.03683 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	4672.47439 MPa
Linear	431 1416
0.2% offset Yield strain	0.01877 mm/mm
0.2% offset Yield stress	76.26 MPa
Toughness	3.68 MJ/mm ³



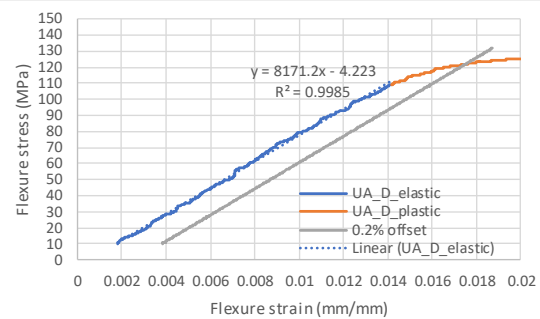
End date	10/16/2022
Specimen label	UA_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26149.05078 N
Flexure extension at Maximum Load	-2.79745 mm
Flexure strain (Extension) at Maximum Load	-0.01759 mm/mm
Flexure stress at Maximum Load	106.15205 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	8390.023403 MPa
Linear	336 1650
0.2% offset Yield strain	0.01526 mm/mm
0.2% offset Yield stress	105.05 MPa
Toughness	6.05 MJ/mm ³



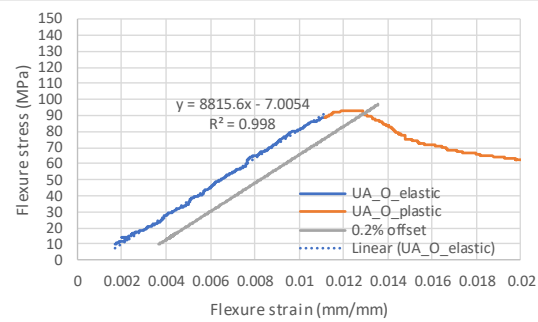
End date	10/16/2022
Specimen label	UA_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	23285.82617 N
Flexure extension at Maximum Load	-4.01521 mm
Flexure strain (Extension) at Maximum Load	-0.02525 mm/mm
Flexure stress at Maximum Load	94.52879 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9795.025319 MPa
Linear	309 1275
0.2% offset Yield strain	0.01008 mm/mm
0.2% offset Yield stress	86.26 MPa
Toughness	4.76 MJ/mm ³



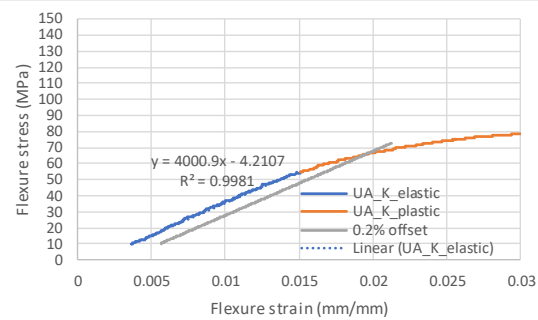
End date	10/16/2022
Specimen label	UA_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	31064.73633 N
Flexure extension at Maximum Load	-3.58217 mm
Flexure strain (Extension) at Maximum Load	-0.02253 mm/mm
Flexure stress at Maximum Load	126.10728 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	8171.153167 MPa
Linear	351 1850
0.2% offset Yield strain	0.01756 mm/mm
0.2% offset Yield stress	121.97 MPa
Toughness	7.26 MJ/mm ³



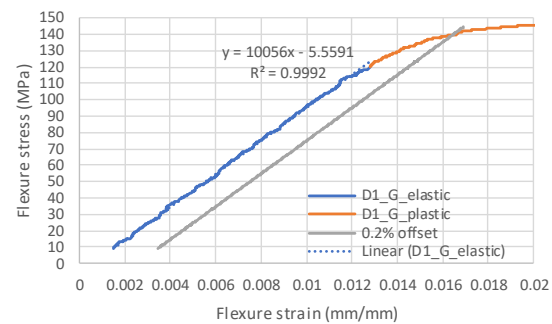
End date	10/16/2022
Specimen label	UA_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	22961.14062 N
Flexure extension at Maximum Load	-1.97597 mm
Flexure strain (Extension) at Maximum Load	-0.01243 mm/mm
Flexure stress at Maximum Load	93.21073 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	8815.623015 MPa
Linear	360 1600
0.2% offset Yield strain	0.01243 mm/mm
0.2% offset Yield stress	93.21 MPa
Toughness	3.54 MJ/mm ³



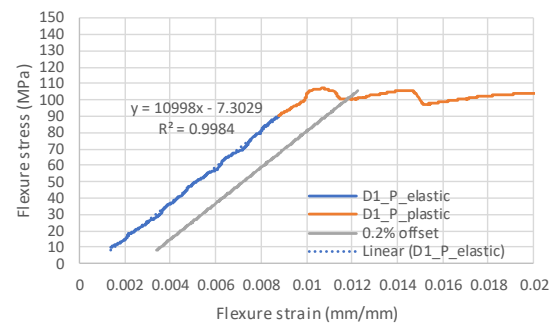
End date	10/16/2022
Specimen label	UA_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	19386.32617 N
Flexure extension at Maximum Load	-5.00921 mm
Flexure strain (Extension) at Maximum Load	-0.0315 mm/mm
Flexure stress at Maximum Load	78.69878 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	4000.917204 MPa
Linear	538 1750
0.2% offset Yield strain	0.01956 mm/mm
0.2% offset Yield stress	66.04 MPa
Toughness	4.42 MJ/mm ³



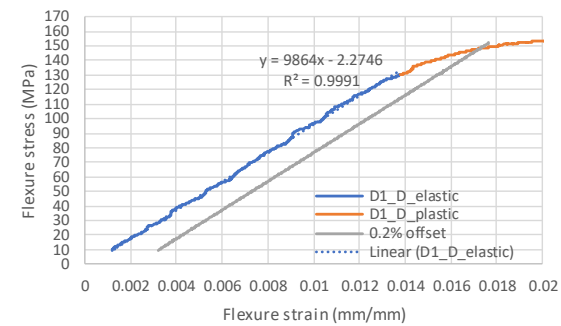
End date	10/15/2022
Specimen label	D1_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	36209.85938 N
Flexure extension at Maximum Load	-3.92292 mm
Flexure strain (Extension) at Maximum Load	-0.02467 mm/mm
Flexure stress at Maximum Load	146.9939 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10055.83058 MPa
Linear	515 1817
0.2% offset Yield strain	0.01657 mm/mm
0.2% offset Yield stress	139.94 MPa
Toughness	3.42 MJ/mm ³



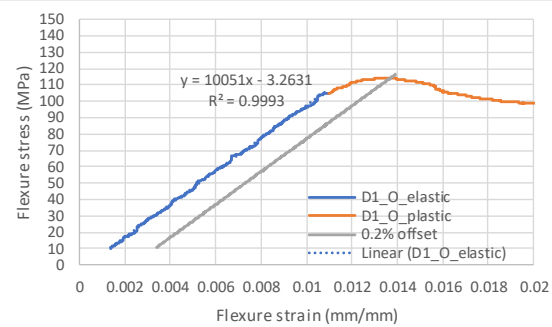
End date	10/15/2022
Specimen label	D1_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26396.91602 N
Flexure extension at Maximum Load	-1.7117 mm
Flexure strain (Extension) at Maximum Load	-0.01076 mm/mm
Flexure stress at Maximum Load	107.15826 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10998.47575 MPa
Linear	497 1334
0.2% offset Yield strain	0.01076 mm/mm
0.2% offset Yield stress	107.16 MPa
Toughness	4.78 MJ/mm ³



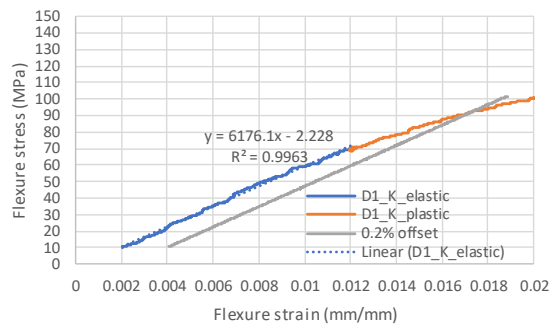
End date	10/15/2022
Specimen label	D1_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	38165.44141 N
Flexure extension at Maximum Load	-3.70784 mm
Flexure strain (Extension) at Maximum Load	-0.02332 mm/mm
Flexure stress at Maximum Load	154.93259 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9864.032112 MPa
Linear	442 1929
0.2% offset Yield strain	0.01737 mm/mm
0.2% offset Yield stress	148.45 MPa
Toughness	7.54 MJ/mm ³



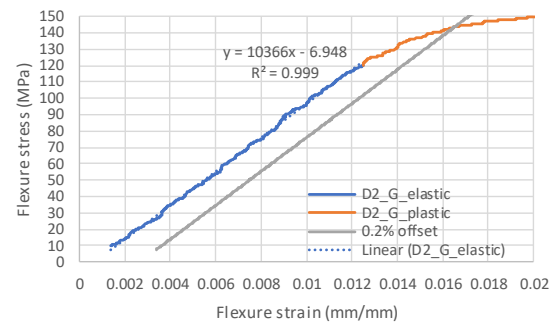
End date	10/15/2022
Specimen label	D1_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28095.21484 N
Flexure extension at Maximum Load	-2.1425 mm
Flexure strain (Extension) at Maximum Load	-0.01347 mm/mm
Flexure stress at Maximum Load	114.05251 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10051.42337 MPa
Linear	465 1649
0.2% offset Yield strain	0.01347 mm/mm
0.2% offset Yield stress	114.05 MPa
Toughness	4.60 MJ/mm ³



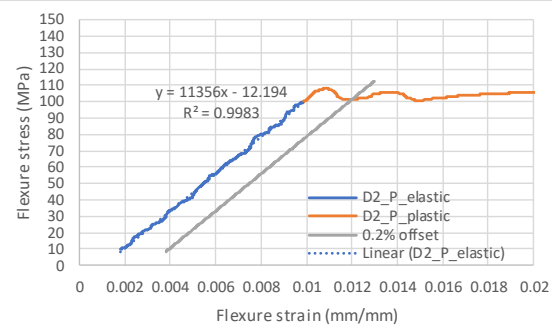
End date	10/15/2022
Specimen label	D1_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28653.68164 N
Flexure extension at Maximum Load	-5.25504 mm
Flexure strain (Extension) at Maximum Load	-0.03305 mm/mm
Flexure stress at Maximum Load	116.3196 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	6176.1416 MPa
Linear	527 1761
0.2% offset Yield strain	0.01718 mm/mm
0.2% offset Yield stress	92.16 MPa
Toughness	4.56 MJ/mm ³



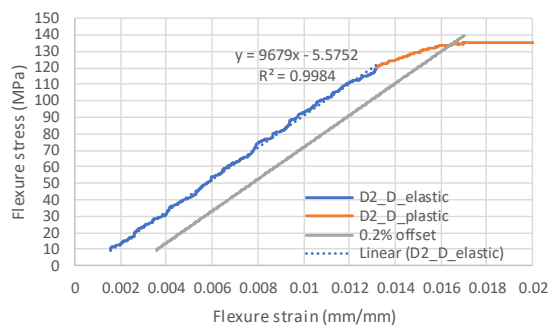
End date	10/15/2022
Specimen label	D2_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	37282.80078 N
Flexure extension at Maximum Load	-3.93488 mm
Flexure strain (Extension) at Maximum Load	-0.02474 mm/mm
Flexure stress at Maximum Load	151.3495 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10366.38585 MPa
Linear	313 1606
0.2% offset Yield strain	0.01654 mm/mm
0.2% offset Yield stress	143.58 MPa
Toughness	4.29 MJ/mm ³



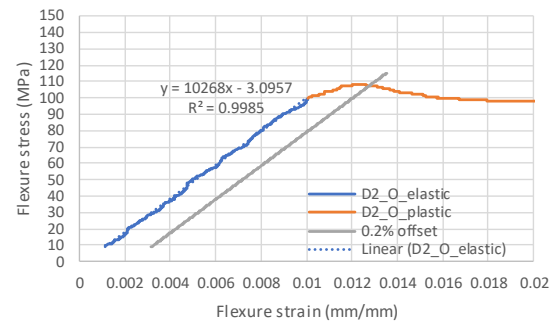
End date	10/15/2022
Specimen label	D2_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26690.68945 N
Flexure extension at Maximum Load	-1.74204 mm
Flexure strain (Extension) at Maximum Load	-0.01095 mm/mm
Flexure stress at Maximum Load	108.35084 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11355.97416 MPa
Linear	577 1486
0.2% offset Yield strain	0.01095 mm/mm
0.2% offset Yield stress	108.35 MPa
Toughness	4.67 MJ/mm ³



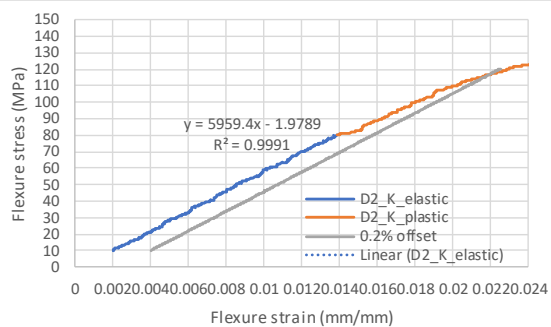
End date	10/15/2022
Specimen label	U2_D was D2_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	33323.94531 N
Flexure extension at Maximum Load	-2.87346 mm
Flexure strain (Extension) at Maximum Load	-0.01807 mm/mm
Flexure stress at Maximum Load	135.27853 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9678.984331 MPa
Linear	440 1845
0.2% offset Yield strain	0.01659 mm/mm
0.2% offset Yield stress	134.36 MPa
Toughness	5.95 MJ/mm ³

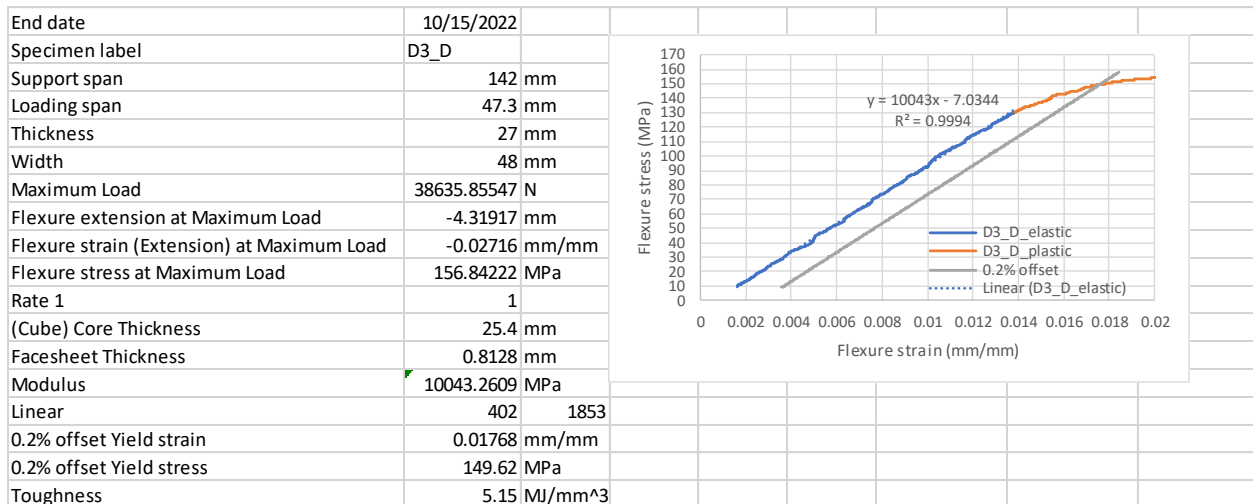
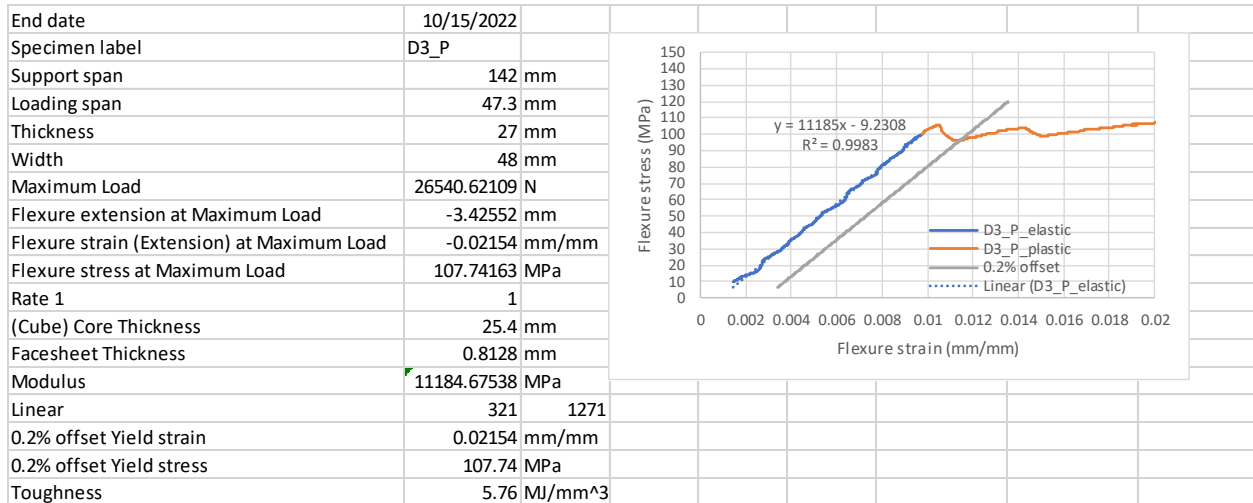
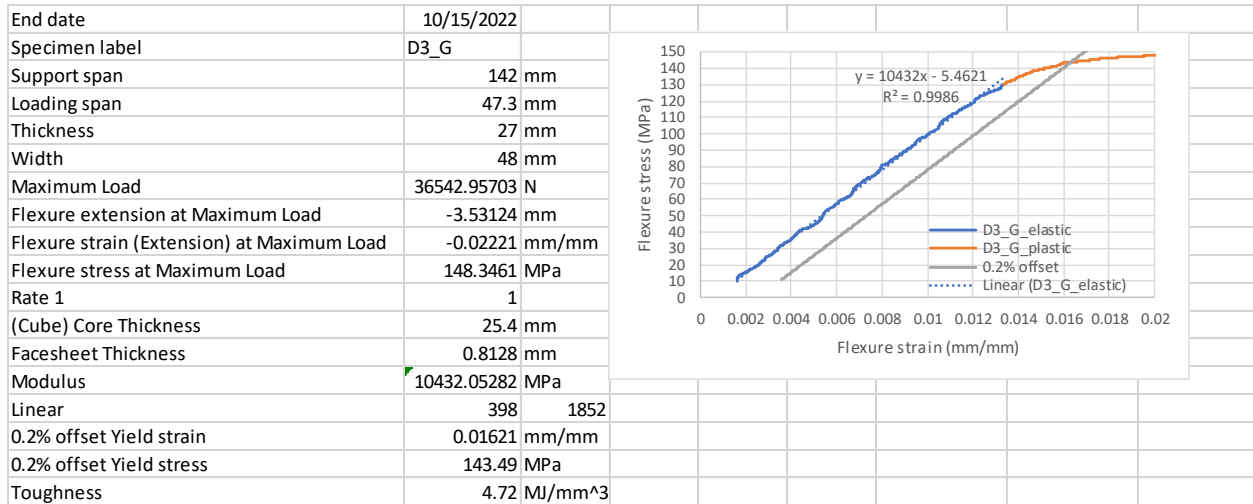


End date	10/15/2022
Specimen label	D2_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26592.53516 N
Flexure extension at Maximum Load	-1.99483 mm
Flexure strain (Extension) at Maximum Load	-0.01254 mm/mm
Flexure stress at Maximum Load	107.95238 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10267.52383 MPa
Linear	467 1593
0.2% offset Yield strain	0.01254 mm/mm
0.2% offset Yield stress	107.95 MPa
Toughness	5.38 MJ/mm ³

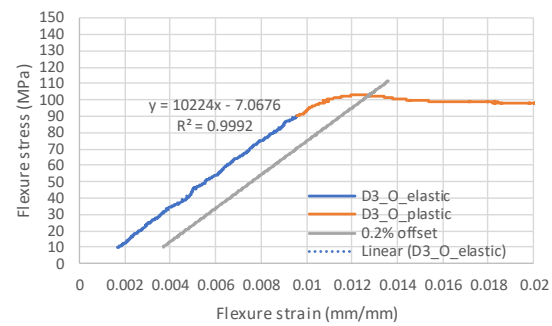


End date	10/15/2022
Specimen label	D2_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	30909.95508 N
Flexure extension at Maximum Load	-4.0303 mm
Flexure strain (Extension) at Maximum Load	-0.02534 mm/mm
Flexure stress at Maximum Load	125.47894 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	5959.410004 MPa
Linear	597 1866
0.2% offset Yield strain	0.02192 mm/mm
0.2% offset Yield stress	116.90 MPa
Toughness	4.01 MJ/mm ³

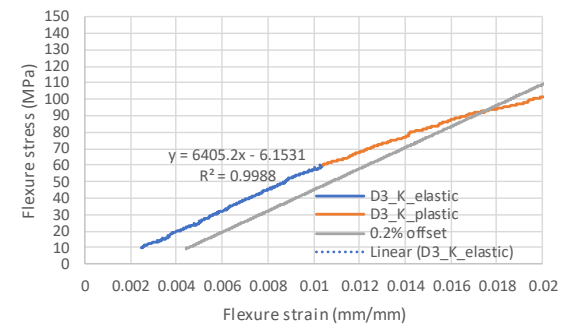




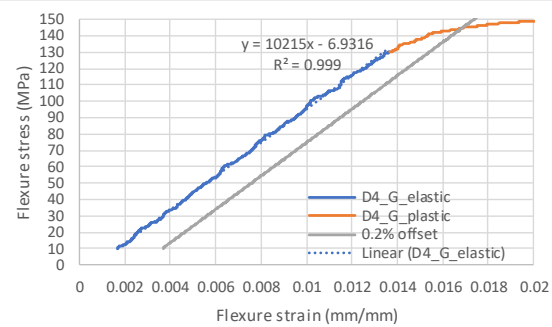
End date	10/15/2022
Specimen label	D3_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	25339.54102 N
Flexure extension at Maximum Load	-1.99419 mm
Flexure strain (Extension) at Maximum Load	-0.01254 mm/mm
Flexure stress at Maximum Load	102.86584 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10223.65124 MPa
Linear	407 1431
0.2% offset Yield strain	0.01254 mm/mm
0.2% offset Yield stress	102.87 MPa
Toughness	5.43 MJ/mm ³



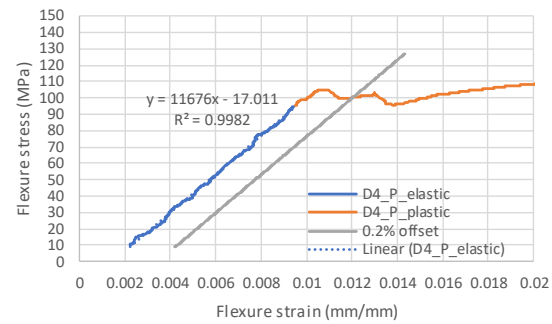
End date	10/15/2022
Specimen label	D3_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	30501.93359 N
Flexure extension at Maximum Load	-5.51972 mm
Flexure strain (Extension) at Maximum Load	-0.03471 mm/mm
Flexure stress at Maximum Load	123.82258 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	6405.163907 MPa
Linear	404 1389
0.2% offset Yield strain	0.01755 mm/mm
0.2% offset Yield stress	92.94 MPa
Toughness	4.45 MJ/mm ³



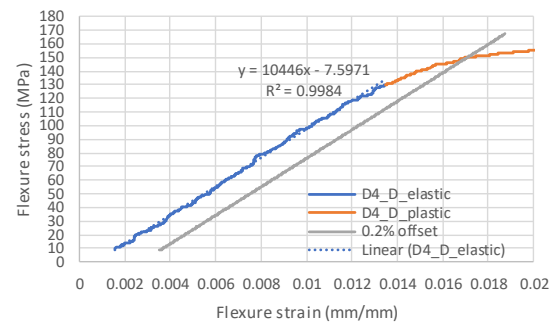
End date	10/16/2022
Specimen label	D4_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	36886.57422 N
Flexure extension at Maximum Load	-3.88298 mm
Flexure strain (Extension) at Maximum Load	-0.02442 mm/mm
Flexure stress at Maximum Load	149.74103 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10214.98681 MPa
Linear	308 1722
0.2% offset Yield strain	0.01676 mm/mm
0.2% offset Yield stress	144.51 MPa
Toughness	3.93 MJ/mm ³



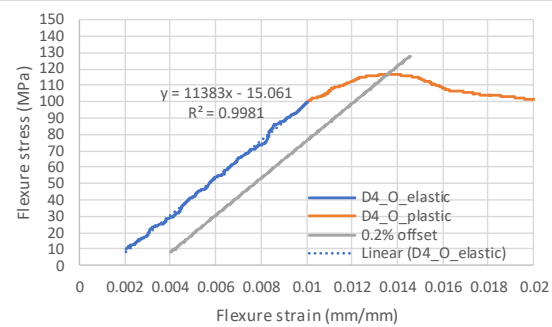
End date	10/16/2022
Specimen label	D4_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26964.11914 N
Flexure extension at Maximum Load	-3.46834 mm
Flexure strain (Extension) at Maximum Load	-0.02181 mm/mm
Flexure stress at Maximum Load	109.46082 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11675.92339 MPa
Linear	315 1180
0.2% offset Yield strain	0.02181 mm/mm
0.2% offset Yield stress	109.46 MPa
Toughness	5.49 MJ/mm ³



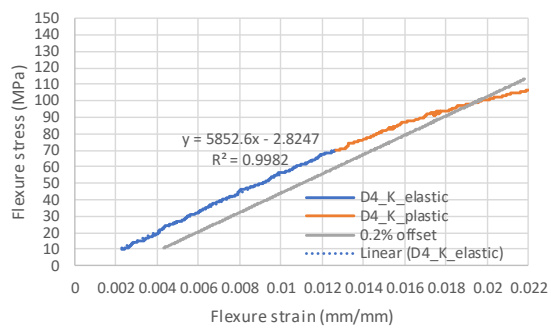
End date	10/16/2022
Specimen label	D4_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	38741.59766 N
Flexure extension at Maximum Load	-3.89885 mm
Flexure strain (Extension) at Maximum Load	-0.02452 mm/mm
Flexure stress at Maximum Load	157.27148 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10446.46267 MPa
Linear	289 1734
0.2% offset Yield strain	0.01704 mm/mm
0.2% offset Yield stress	149.78 MPa
Toughness	7.18 MJ/mm ³



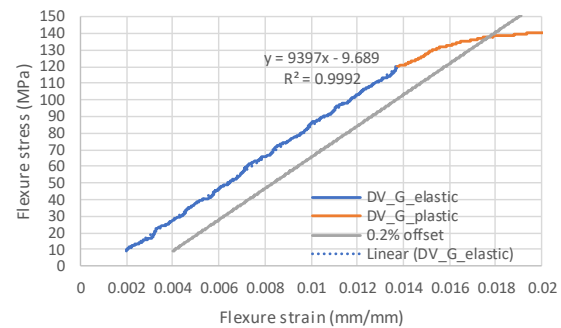
End date	10/16/2022
Specimen label	D4_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28753.08008 N
Flexure extension at Maximum Load	-2.16393 mm
Flexure strain (Extension) at Maximum Load	-0.01361 mm/mm
Flexure stress at Maximum Load	116.72311 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11382.92513 MPa
Linear	362 1446
0.2% offset Yield strain	0.01361 mm/mm
0.2% offset Yield stress	116.72 MPa
Toughness	5.69 MJ/mm ³



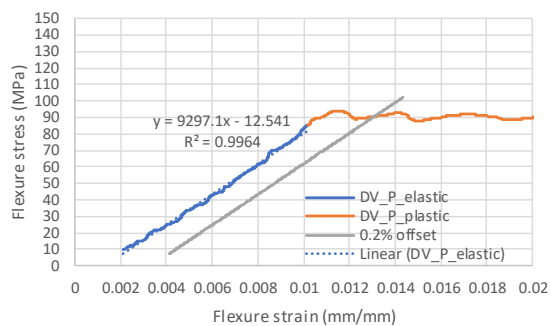
End date	10/16/2022
Specimen label	D4_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	30774.41016 N
Flexure extension at Maximum Load	-5.85919 mm
Flexure strain (Extension) at Maximum Load	-0.03684 mm/mm
Flexure stress at Maximum Load	124.9287 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	5852.625734 MPa
Linear	430 1670
0.2% offset Yield strain	0.01951 mm/mm
0.2% offset Yield stress	99.02 MPa
Toughness	6.15 MJ/mm ³



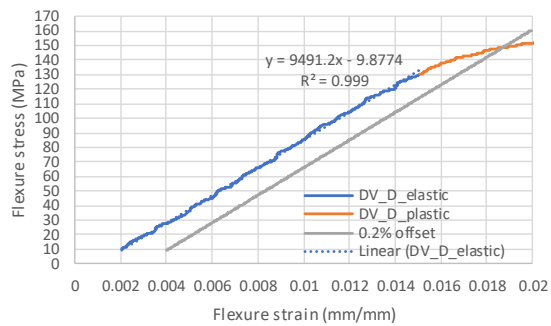
End date	10/16/2022
Specimen label	DV_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	34827.8125 N
Flexure extension at Maximum Load	-2.83069 mm
Flexure strain (Extension) at Maximum Load	-0.0178 mm/mm
Flexure stress at Maximum Load	141.38348 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9397.02295 MPa
Linear	311 1702
0.2% offset Yield strain	0.01777 mm/mm
0.2% offset Yield stress	138.15 MPa
Toughness	3.71 MJ/mm ³



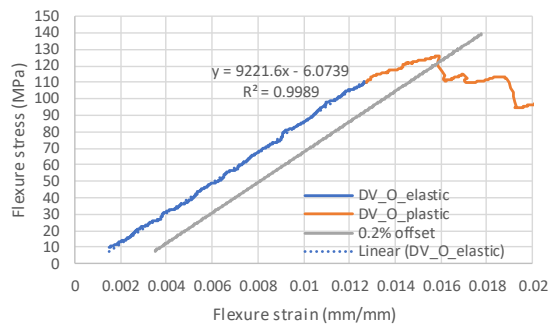
End date	10/16/2022
Specimen label	DV_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	23293.29688 N
Flexure extension at Maximum Load	-3.97694 mm
Flexure strain (Extension) at Maximum Load	-0.02501 mm/mm
Flexure stress at Maximum Load	94.55912 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9297.131337 MPa
Linear	451 1335
0.2% offset Yield strain	0.02501 mm/mm
0.2% offset Yield stress	94.56 MPa
Toughness	3.10 MJ/mm ³



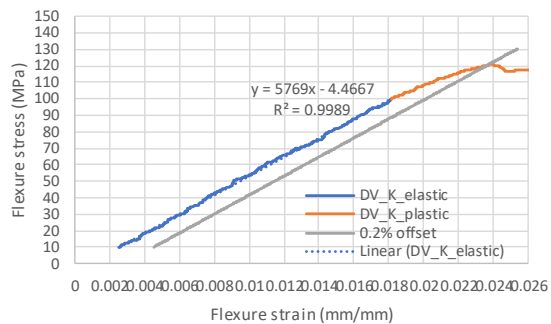
End date	10/16/2022
Specimen label	DV_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	39356.28516 N
Flexure extension at Maximum Load	-4.72041 mm
Flexure strain (Extension) at Maximum Load	-0.02968 mm/mm
Flexure stress at Maximum Load	159.76682 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9491.180491 MPa
Linear	298 1798
0.2% offset Yield strain	0.0186 mm/mm
0.2% offset Yield stress	148.56 MPa
Toughness	4.85 MJ/mm ³



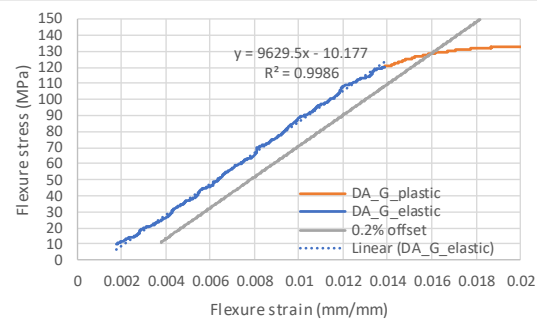
End date	10/16/2022
Specimen label	DV_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	31039.28906 N
Flexure extension at Maximum Load	-2.52842 mm
Flexure strain (Extension) at Maximum Load	-0.0159 mm/mm
Flexure stress at Maximum Load	126.00397 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9221.646326 MPa
Linear	282 1493
0.2% offset Yield strain	0.0159 mm/mm
0.2% offset Yield stress	126.00 MPa
Toughness	2.01 MJ/mm ³



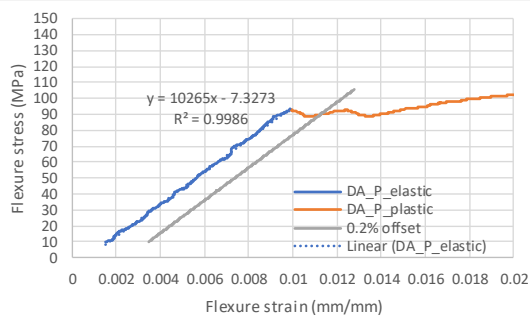
End date	10/16/2022
Specimen label	DV_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	29720.94922 N
Flexure extension at Maximum Load	-3.80813 mm
Flexure strain (Extension) at Maximum Load	-0.02395 mm/mm
Flexure stress at Maximum Load	120.65217 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	5769.049578 MPa
Linear	354 2029
0.2% offset Yield strain	0.02375 mm/mm
0.2% offset Yield stress	120.55 MPa
Toughness	3.19 MJ/mm ³



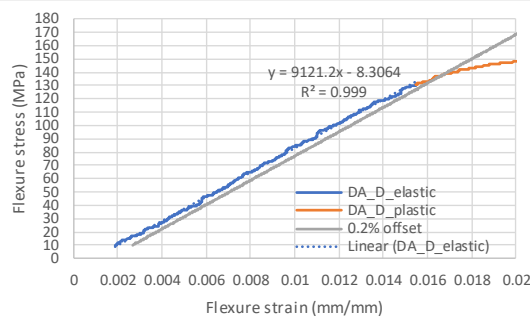
End date	10/16/2022
Specimen label	DA_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	33042.03516 N
Flexure extension at Maximum Load	-5.80003 mm
Flexure strain (Extension) at Maximum Load	-0.03647 mm/mm
Flexure stress at Maximum Load	134.13412 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9629.504252 MPa
Linear	346 1800
0.2% offset Yield strain	0.01679 mm/mm
0.2% offset Yield stress	129.93 MPa
Toughness	5.90 MJ/mm ³



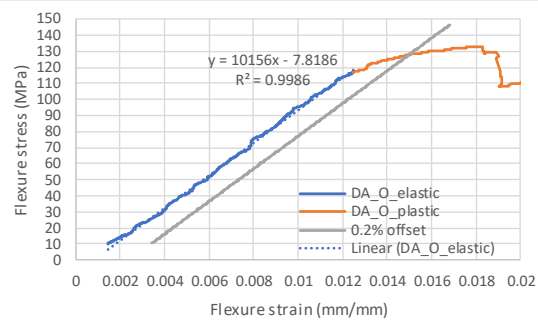
End date	10/16/2022
Specimen label	DA_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	26149.80664 N
Flexure extension at Maximum Load	-4.22755 mm
Flexure strain (Extension) at Maximum Load	-0.02658 mm/mm
Flexure stress at Maximum Load	106.15512 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10265.23653 MPa
Linear	314 1250
0.2% offset Yield strain	0.01189 mm/mm
0.2% offset Yield stress	96.52 MPa
Toughness	5.42 MJ/mm ³



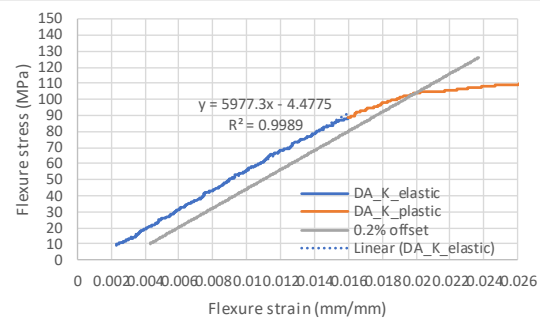
End date	10/16/2022
Specimen label	DA_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	38086.82422 N
Flexure extension at Maximum Load	-4.75365 mm
Flexure strain (Extension) at Maximum Load	-0.02989 mm/mm
Flexure stress at Maximum Load	154.61343 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	9121.181575 MPa
Linear	449 2050
0.2% offset Yield strain	0.017 mm/mm
0.2% offset Yield stress	139.16 MPa
Toughness	7.98 MJ/mm ³



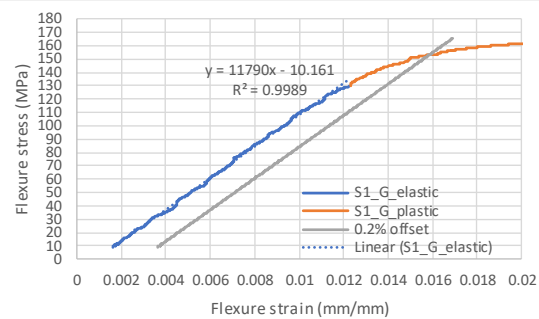
End date	10/16/2022
Specimen label	DA_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	32683.0332 N
Flexure extension at Maximum Load	-2.8926 mm
Flexure strain (Extension) at Maximum Load	-0.01819 mm/mm
Flexure stress at Maximum Load	132.67674 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10156.1387 MPa
Linear	292 1550
0.2% offset Yield strain	0.01525 mm/mm
0.2% offset Yield stress	128.32 MPa
Toughness	3.28 MJ/mm ³



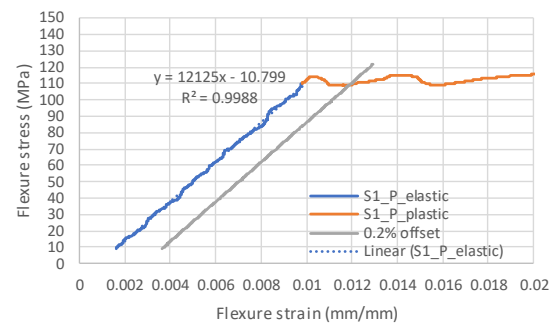
End date	10/16/2022
Specimen label	DA_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	29445.99609 N
Flexure extension at Maximum Load	-8.0895 mm
Flexure strain (Extension) at Maximum Load	-0.05087 mm/mm
Flexure stress at Maximum Load	119.536 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	5977.309237 MPa
Linear	407 1950
0.2% offset Yield strain	0.01982 mm/mm
0.2% offset Yield stress	103.60 MPa
Toughness	6.90 MJ/mm ³



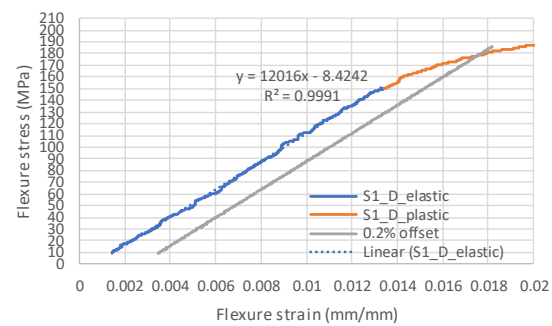
End date	10/15/2022
Specimen label	S1_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	40525.625 N
Flexure extension at Maximum Load	-3.84284 mm
Flexure strain (Extension) at Maximum Load	-0.02417 mm/mm
Flexure stress at Maximum Load	164.51375 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11789.74042 MPa
Linear	295 1591
0.2% offset Yield strain	0.01575 mm/mm
0.2% offset Yield stress	152.82 MPa
Toughness	3.07 MJ/mm ³



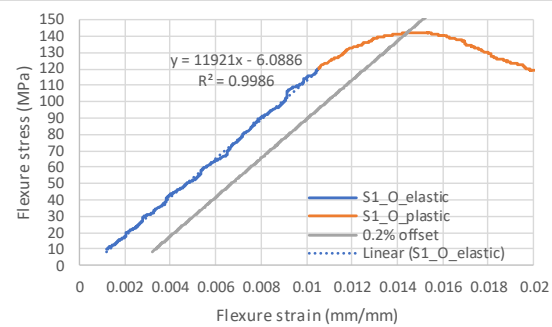
End date	10/15/2022
Specimen label	S1_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28601.09961 N
Flexure extension at Maximum Load	-3.39851 mm
Flexure strain (Extension) at Maximum Load	-0.02137 mm/mm
Flexure stress at Maximum Load	116.10614 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	12124.9729 MPa
Linear	441 1403
0.2% offset Yield strain	0.02137 mm/mm
0.2% offset Yield stress	116.11 MPa
Toughness	5.89 MJ/mm ³



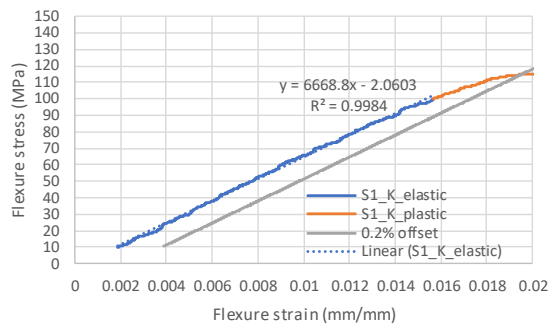
End date	10/15/2022
Specimen label	S1_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	48700.04688 N
Flexure extension at Maximum Load	-4.6336 mm
Flexure strain (Extension) at Maximum Load	-0.02914 mm/mm
Flexure stress at Maximum Load	197.6978 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	12015.71893 MPa
Linear	362 1771
0.2% offset Yield strain	0.01758 mm/mm
0.2% offset Yield stress	178.44 MPa
Toughness	4.76 MJ/mm ³



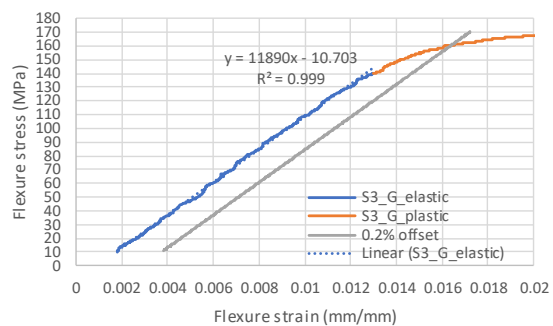
End date	10/15/2022
Specimen label	S1_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	35106.95312 N
Flexure extension at Maximum Load	-2.31523 mm
Flexure strain (Extension) at Maximum Load	-0.01456 mm/mm
Flexure stress at Maximum Load	142.51665 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11920.75002 MPa
Linear	416 1557
0.2% offset Yield strain	0.01456 mm/mm
0.2% offset Yield stress	142.52 MPa
Toughness	6.26 MJ/mm ³



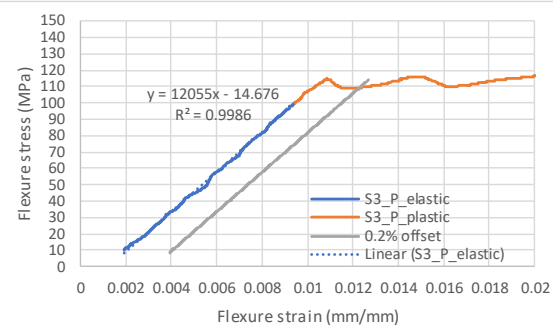
End date	10/15/2022
Specimen label	S1_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28714.33008 N
Flexure extension at Maximum Load	-4.15536 mm
Flexure strain (Extension) at Maximum Load	-0.02613 mm/mm
Flexure stress at Maximum Load	116.5658 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	6668.79009 MPa
Linear	359 1931
0.2% offset Yield strain	0.0196 mm/mm
0.2% offset Yield stress	114.67 MPa
Toughness	3.27 MJ/mm ³



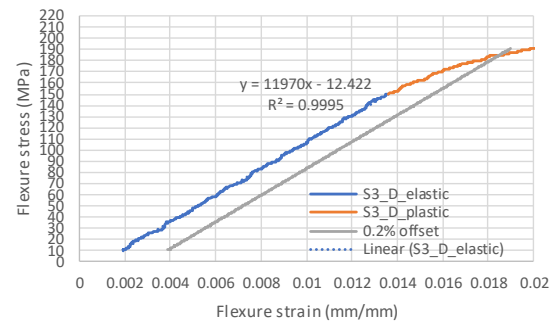
End date	10/15/2022
Specimen label	S3_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	41899.9375 N
Flexure extension at Maximum Load	-3.79403 mm
Flexure strain (Extension) at Maximum Load	-0.02386 mm/mm
Flexure stress at Maximum Load	170.09276 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11890.186 MPa
Linear	238 1629
0.2% offset Yield strain	0.01629 mm/mm
0.2% offset Yield stress	160.53 MPa
Toughness	2.76 MJ/mm ³



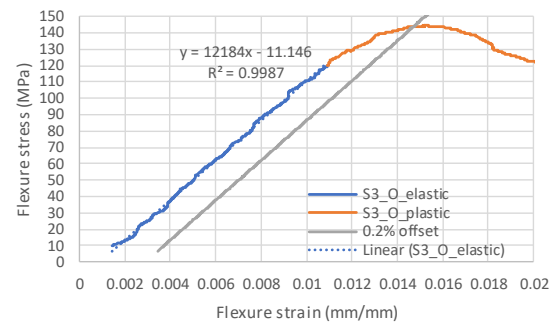
End date	10/15/2022
Specimen label	S3_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28914.55273 N
Flexure extension at Maximum Load	-3.44915 mm
Flexure strain (Extension) at Maximum Load	-0.02169 mm/mm
Flexure stress at Maximum Load	117.37861 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	12055.13985 MPa
Linear	290 1180
0.2% offset Yield strain	0.02169 mm/mm
0.2% offset Yield stress	117.38 MPa
Toughness	5.83 MJ/mm ³



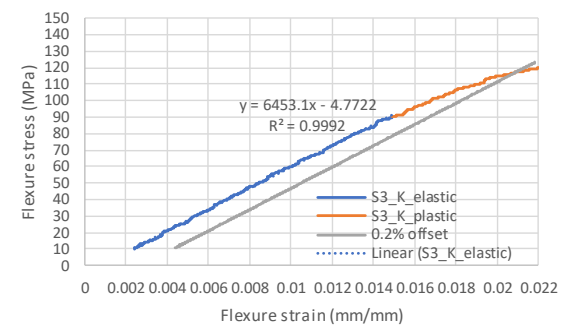
End date	10/15/2022
Specimen label	S3_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	50102.42969 N
Flexure extension at Maximum Load	-4.68067 mm
Flexure strain (Extension) at Maximum Load	-0.02943 mm/mm
Flexure stress at Maximum Load	203.39078 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11970.29737 MPa
Linear	286 1706
0.2% offset Yield strain	0.01846 mm/mm
0.2% offset Yield stress	185.22 MPa
Toughness	4.91 MJ/mm ³



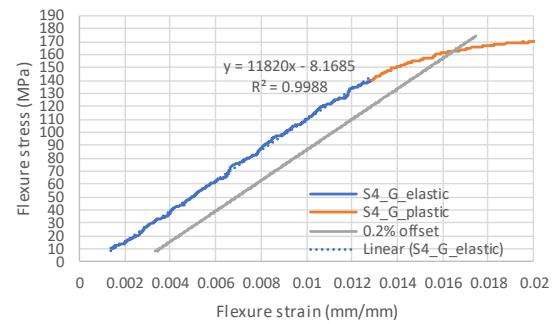
End date	10/15/2022
Specimen label	S3_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	35529.07422 N
Flexure extension at Maximum Load	-2.42165 mm
Flexure strain (Extension) at Maximum Load	-0.01523 mm/mm
Flexure stress at Maximum Load	144.23026 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	12183.92287 MPa
Linear	266 1402
0.2% offset Yield strain	0.0148 mm/mm
0.2% offset Yield stress	143.89 MPa
Toughness	6.42 MJ/mm ³



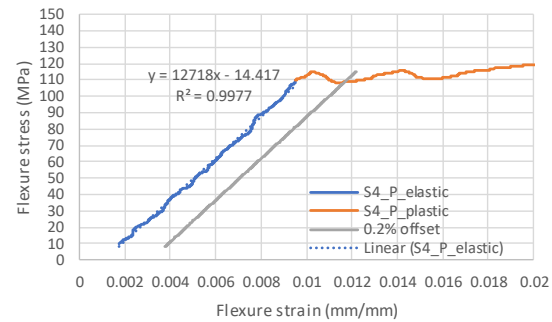
End date	10/15/2022
Specimen label	S3_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	30096.72852 N
Flexure extension at Maximum Load	-4.0891 mm
Flexure strain (Extension) at Maximum Load	-0.02571 mm/mm
Flexure stress at Maximum Load	122.17765 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	6453.144219 MPa
Linear	347 1837
0.2% offset Yield strain	0.02069 mm/mm
0.2% offset Yield stress	116.39 MPa
Toughness	3.19 MJ/mm ³



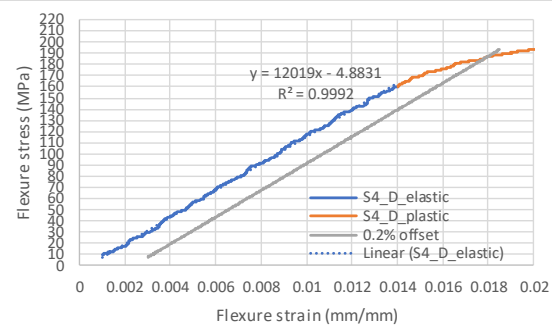
End date	10/16/2022
Specimen label	S4_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	42746.19141 N
Flexure extension at Maximum Load	-4.12146 mm
Flexure strain (Extension) at Maximum Load	-0.02592 mm/mm
Flexure stress at Maximum Load	173.52814 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	11819.87067 MPa
Linear	270 1617
0.2% offset Yield strain	0.01644 mm/mm
0.2% offset Yield stress	162.97 MPa
Toughness	3.30 MJ/mm ³



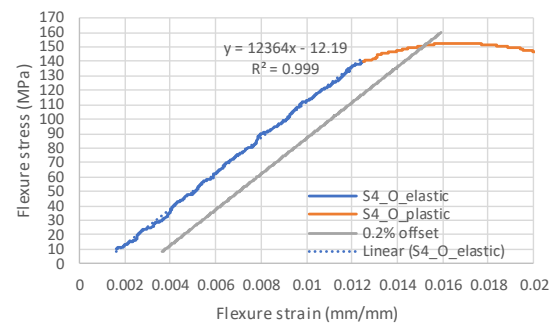
End date	10/16/2022
Specimen label	S4_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	29788.95703 N
Flexure extension at Maximum Load	-3.67086 mm
Flexure strain (Extension) at Maximum Load	-0.02308 mm/mm
Flexure stress at Maximum Load	120.92825 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	12718.26497 MPa
Linear	304 1248
0.2% offset Yield strain	0.02308 mm/mm
0.2% offset Yield stress	120.93 MPa
Toughness	5.94 MJ/mm ³



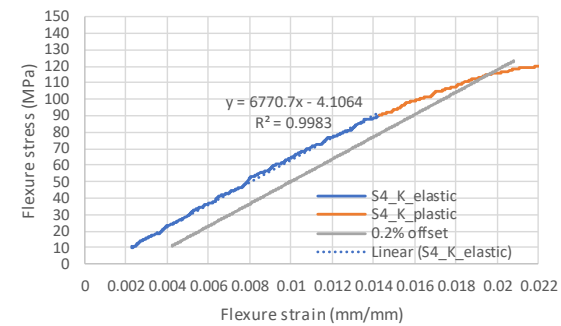
End date	10/16/2022
Specimen label	S4_D
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	50330.30859 N
Flexure extension at Maximum Load	-5.26965 mm
Flexure strain (Extension) at Maximum Load	-0.03314 mm/mm
Flexure stress at Maximum Load	204.31586 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	12019.06028 MPa
Linear	300 1808
0.2% offset Yield strain	0.01791 mm/mm
0.2% offset Yield stress	186.44 MPa
Toughness	5.19 MJ/mm ³



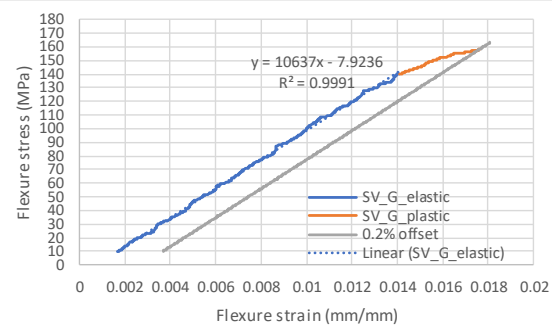
End date	10/16/2022
Specimen label	S4_O
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	37575.32812 N
Flexure extension at Maximum Load	-2.67794 mm
Flexure strain (Extension) at Maximum Load	-0.01684 mm/mm
Flexure stress at Maximum Load	152.53702 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	12363.79018 MPa
Linear	289 1644
0.2% offset Yield strain	0.01525 mm/mm
0.2% offset Yield stress	151.00 MPa
Toughness	6.92 MJ/mm ³

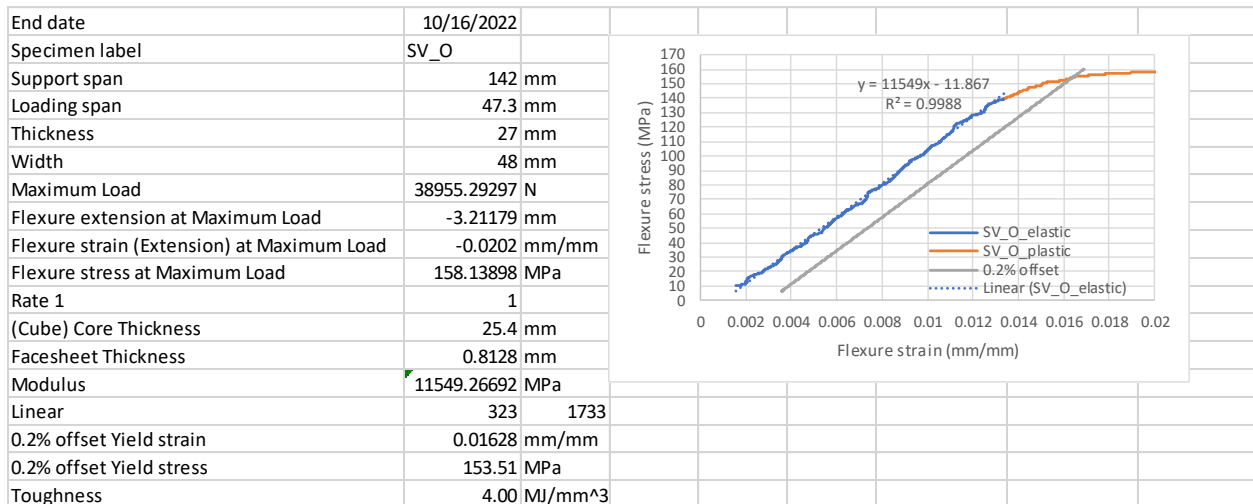
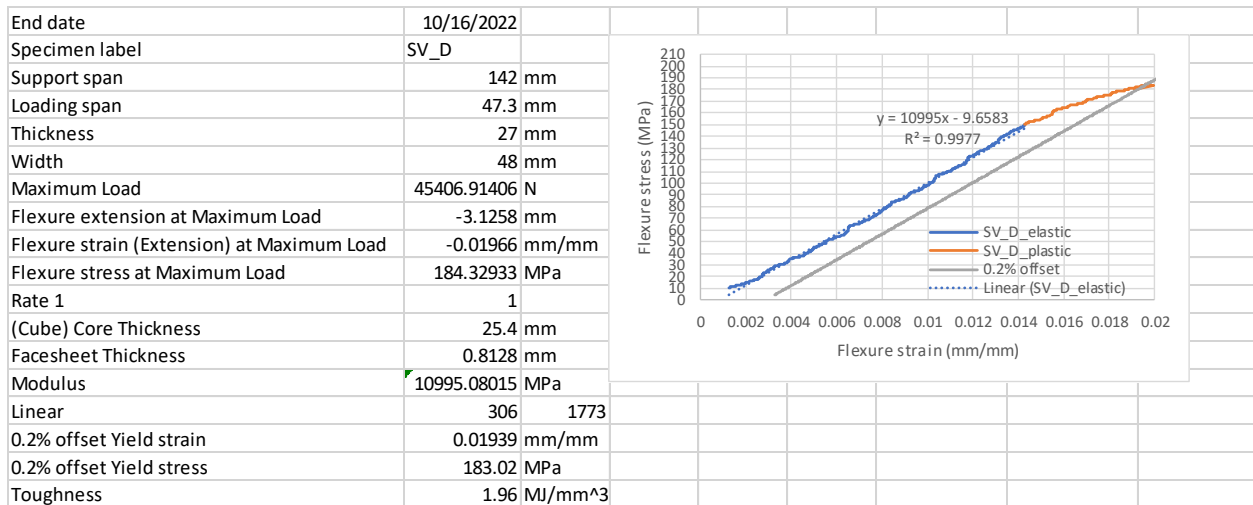
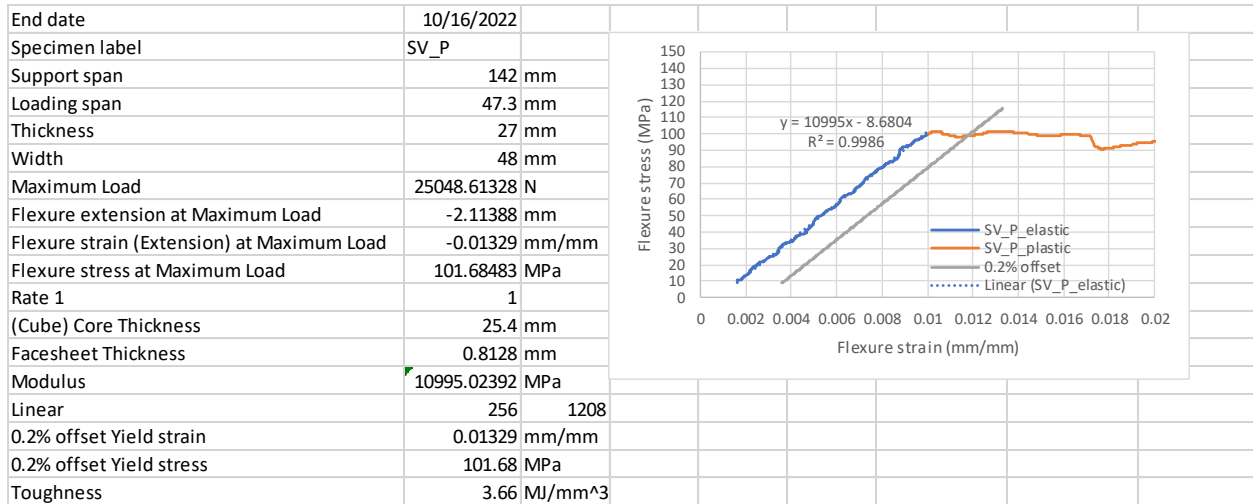


End date	10/16/2022
Specimen label	S4_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	29932.39258 N
Flexure extension at Maximum Load	-3.9141 mm
Flexure strain (Extension) at Maximum Load	-0.02461 mm/mm
Flexure stress at Maximum Load	121.51053 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	6770.675887 MPa
Linear	367 1791
0.2% offset Yield strain	0.01974 mm/mm
0.2% offset Yield stress	115.40 MPa
Toughness	3.84 MJ/mm ³

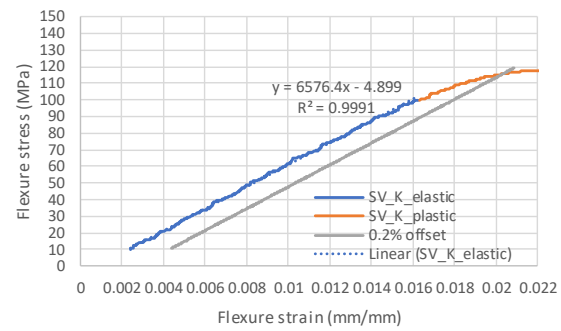


End date	10/16/2022
Specimen label	SV_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	38963.19141 N
Flexure extension at Maximum Load	-2.80091 mm
Flexure strain (Extension) at Maximum Load	-0.01761 mm/mm
Flexure stress at Maximum Load	158.17105 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10636.658 MPa
Linear	369 1790
0.2% offset Yield strain	0.01745 mm/mm
0.2% offset Yield stress	157.39 MPa
Toughness	1.48 MJ/mm ³

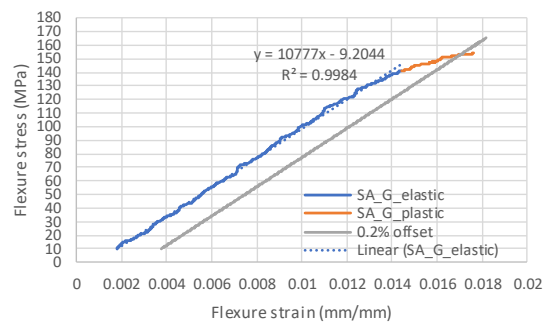




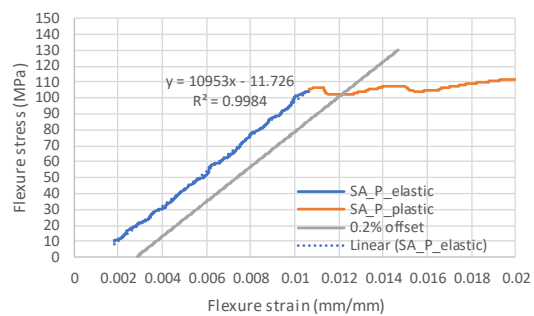
End date	10/16/2022
Specimen label	SV_K
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28957.8125 N
Flexure extension at Maximum Load	-3.49494 mm
Flexure strain (Extension) at Maximum Load	-0.02198 mm/mm
Flexure stress at Maximum Load	117.55422 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	6576.43221 MPa
Linear	361 2001
0.2% offset Yield strain	0.02038 mm/mm
0.2% offset Yield stress	115.62 MPa
Toughness	2.67 MJ/mm ³



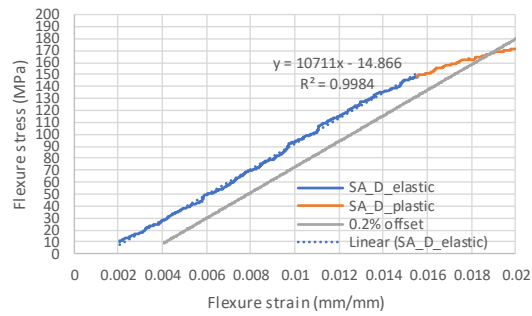
End date	10/16/2022
Specimen label	SA_G
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	39827.54688 N
Flexure extension at Maximum Load	-3.94983 mm
Flexure strain (Extension) at Maximum Load	-0.02484 mm/mm
Flexure stress at Maximum Load	161.6799 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10776.53894 MPa
Linear	332 1800
0.2% offset Yield strain	0.01724 mm/mm
0.2% offset Yield stress	153.08 MPa
Toughness	2.80 MJ/mm ³



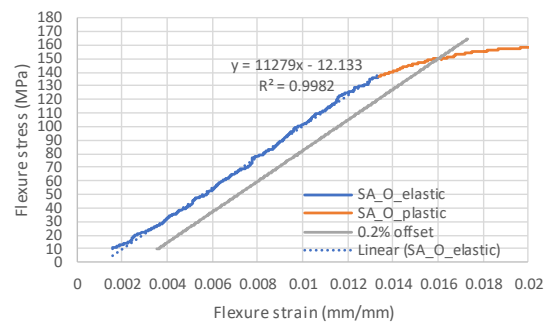
End date	10/16/2022
Specimen label	SA_P
Support span	142 mm
Loading span	47.3 mm
Thickness	27 mm
Width	48 mm
Maximum Load	28624.66016 N
Flexure extension at Maximum Load	-4.40149 mm
Flexure strain (Extension) at Maximum Load	-0.02768 mm/mm
Flexure stress at Maximum Load	116.20179 MPa
Rate 1	1
(Cube) Core Thickness	25.4 mm
Facesheet Thickness	0.8128 mm
Modulus	10953.13929 MPa
Linear	362 1350
0.2% offset Yield strain	0.01106 mm/mm
0.2% offset Yield stress	106.53 MPa
Toughness	4.24 MJ/mm ³



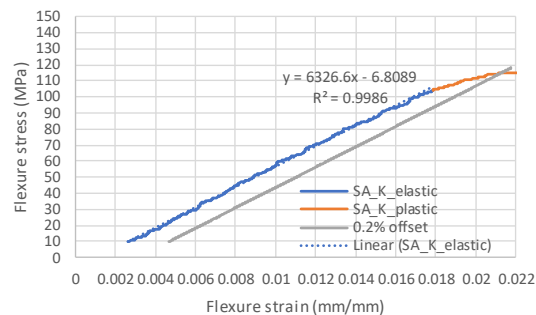
End date	10/16/2022	
Specimen label	SA_D	
Support span	142 mm	
Loading span	47.3 mm	
Thickness	27 mm	
Width	48 mm	
Maximum Load	46917.95312 N	
Flexure extension at Maximum Load	-6.29689 mm	
Flexure strain (Extension) at Maximum Load	-0.0396 mm/mm	
Flexure stress at Maximum Load	190.46339 MPa	
Rate 1	1	
(Cube) Core Thickness	25.4 mm	
Facesheet Thickness	0.8128 mm	
Modulus	10710.60205 MPa	
Linear	363	1850
0.2% offset Yield strain	0.01895 mm/mm	
0.2% offset Yield stress	168.21 MPa	
Toughness	6.23 MJ/mm ³	



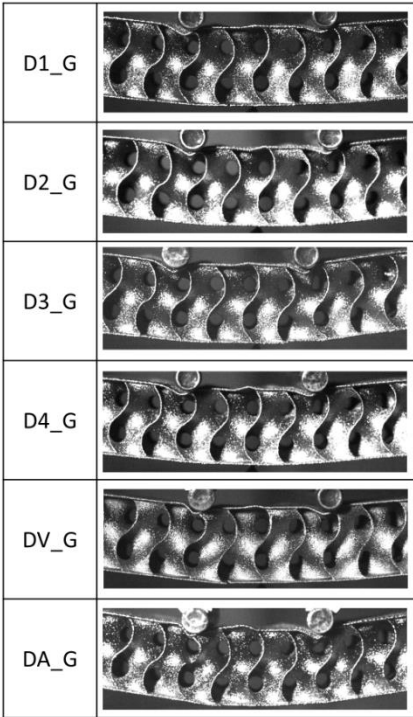
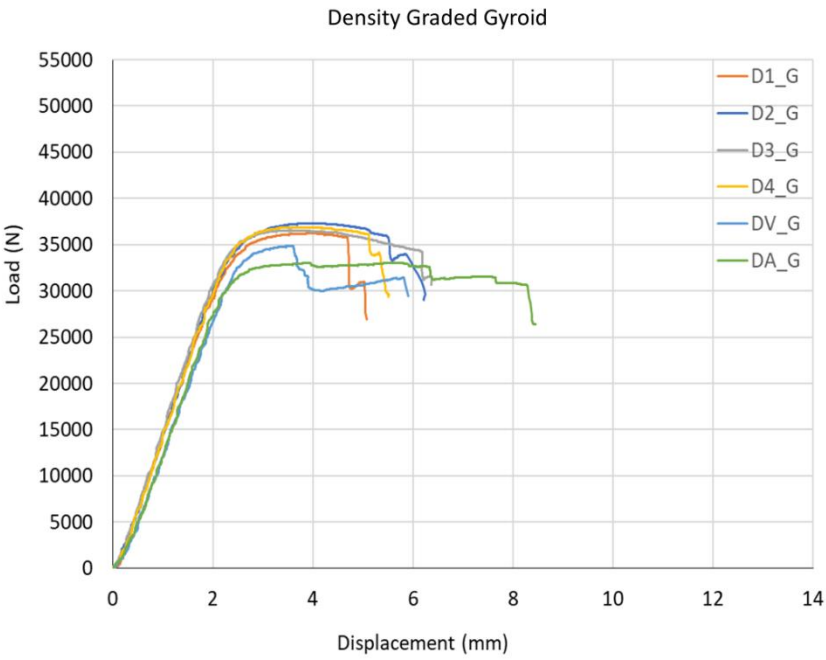
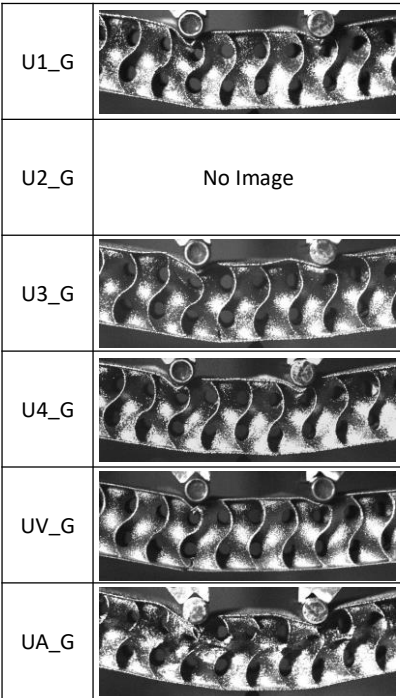
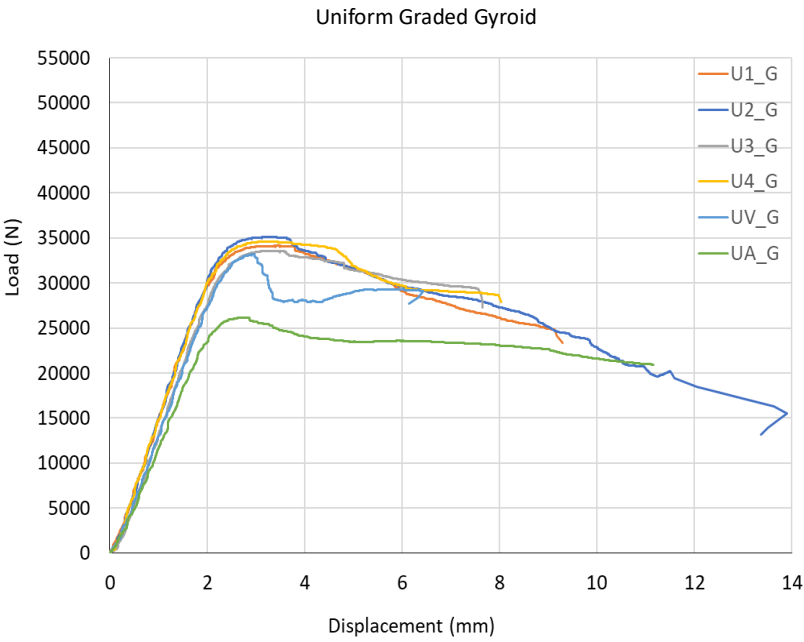
End date	10/16/2022	
Specimen label	SA_O	
Support span	142 mm	
Loading span	47.3 mm	
Thickness	27 mm	
Width	48 mm	
Maximum Load	39065.17188 N	
Flexure extension at Maximum Load	-3.39657 mm	
Flexure strain (Extension) at Maximum Load	-0.02136 mm/mm	
Flexure stress at Maximum Load	158.58504 MPa	
Rate 1	1	
(Cube) Core Thickness	25.4 mm	
Facesheet Thickness	0.8128 mm	
Modulus	11279.03406 MPa	
Linear	334	1700
0.2% offset Yield strain	0.01603 mm/mm	
0.2% offset Yield stress	150.11 MPa	
Toughness	7.70 MJ/mm ³	

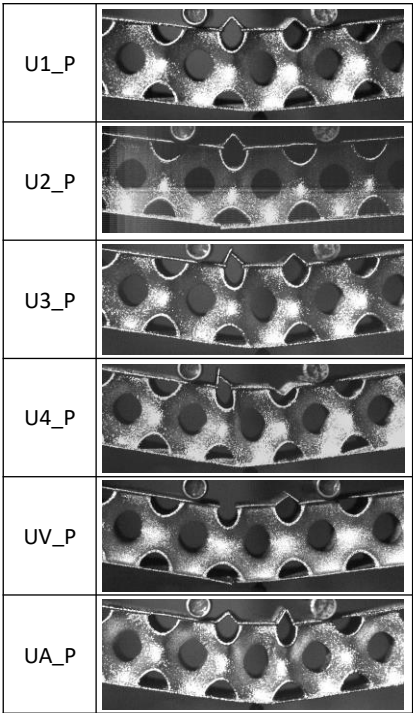
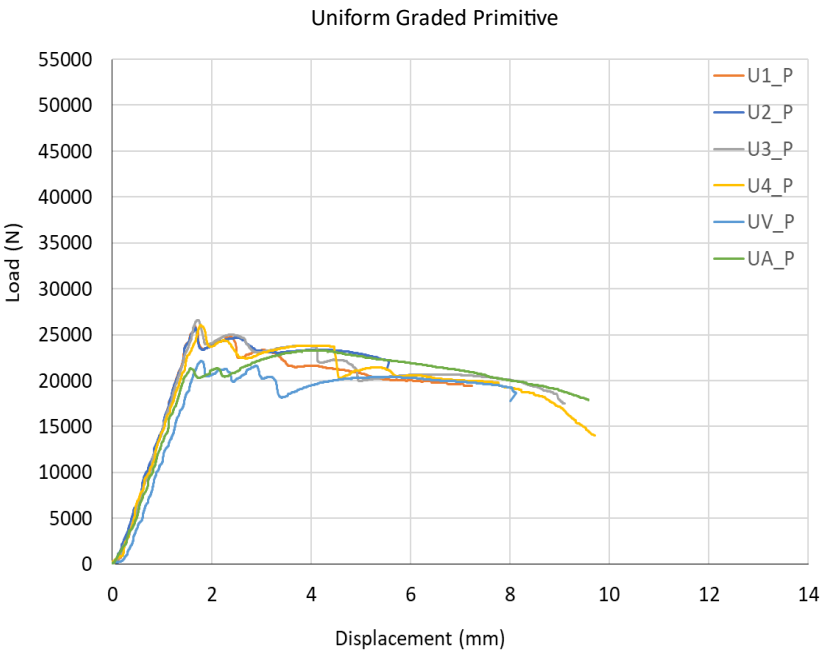
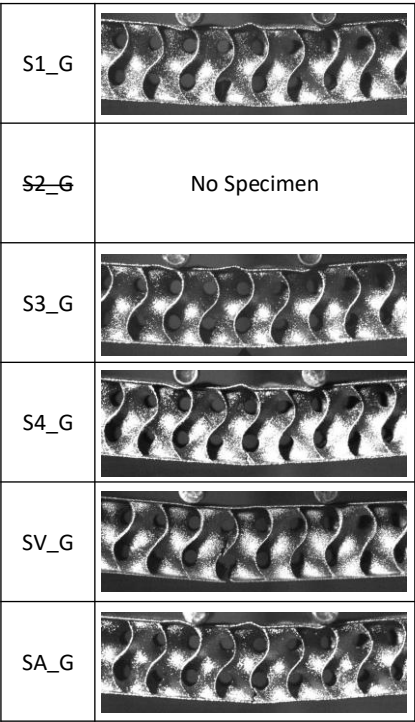
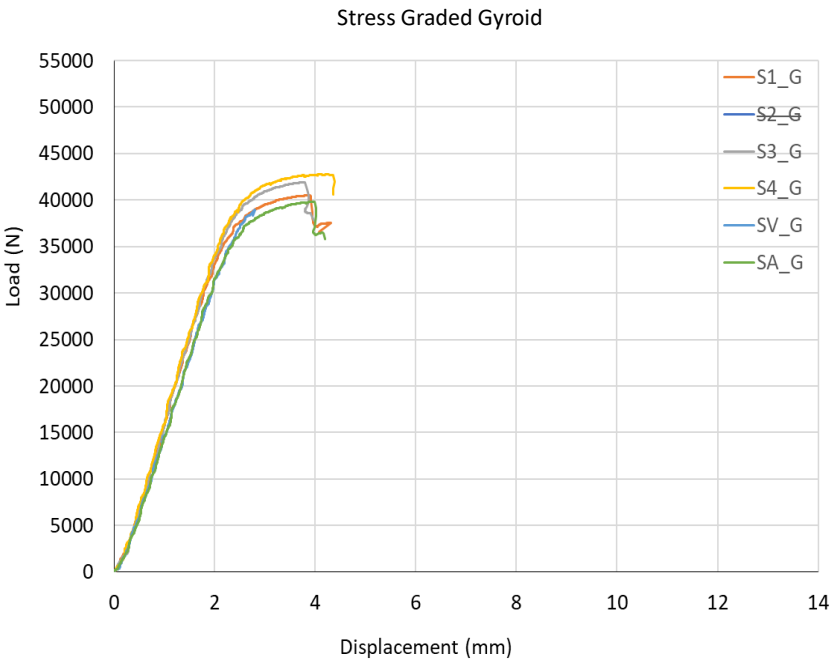


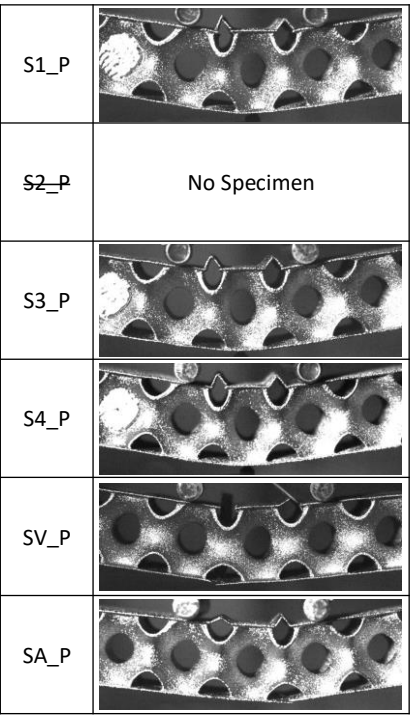
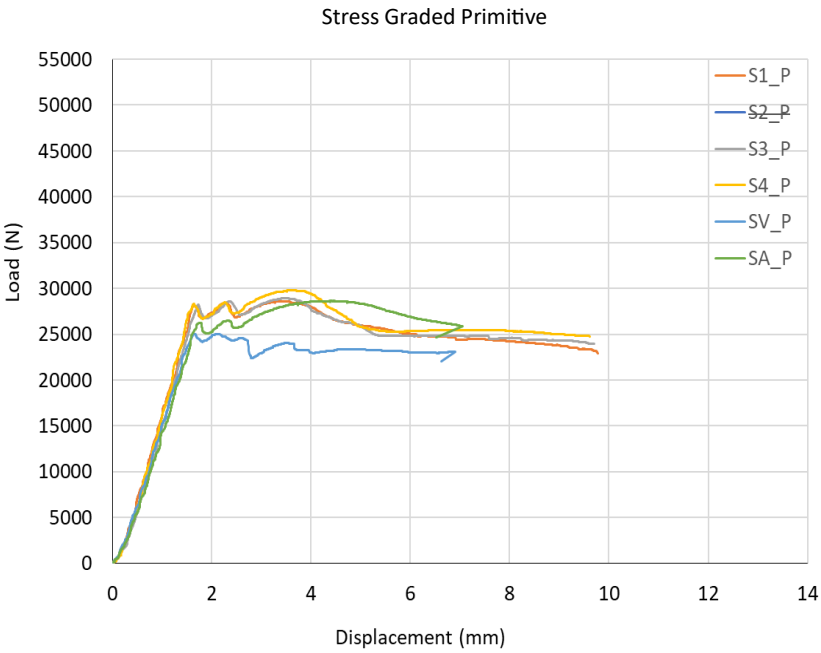
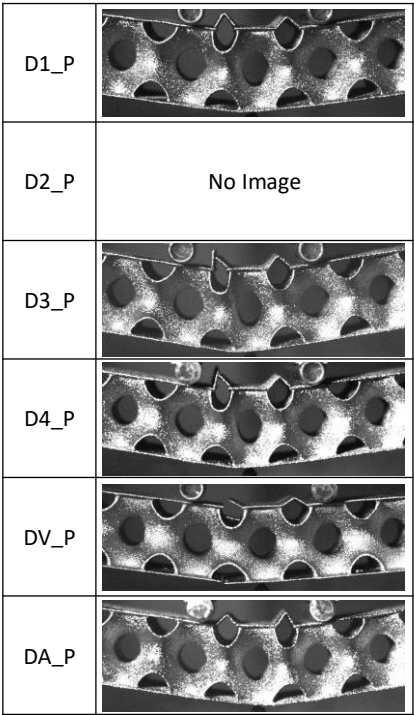
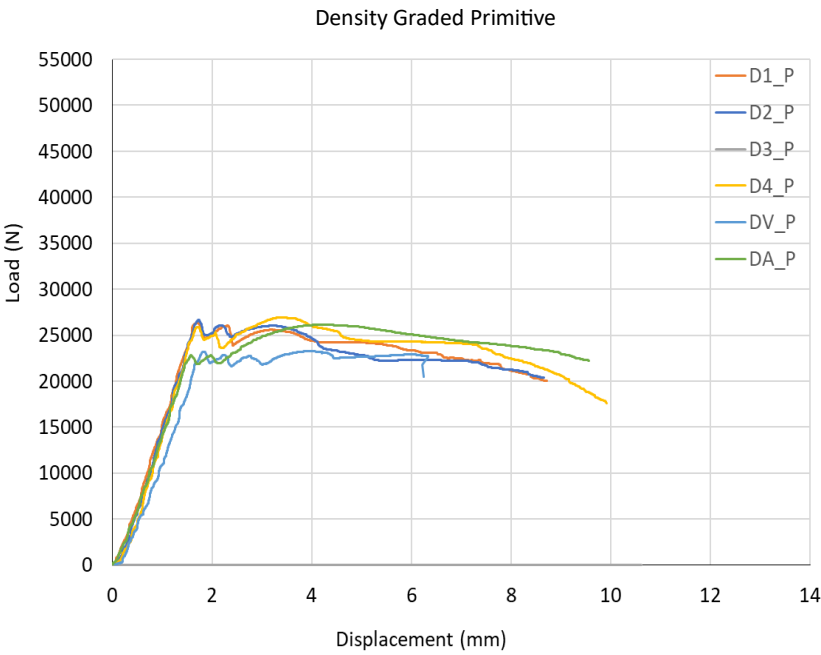
End date	10/16/2022	
Specimen label	SA_K	
Support span	142 mm	
Loading span	47.3 mm	
Thickness	27 mm	
Width	48 mm	
Maximum Load	28383.45898 N	
Flexure extension at Maximum Load	-4.56006 mm	
Flexure strain (Extension) at Maximum Load	-0.02868 mm/mm	
Flexure stress at Maximum Load	115.22263 MPa	
Rate 1	1	
(Cube) Core Thickness	25.4 mm	
Facesheet Thickness	0.8128 mm	
Modulus	6326.617191 MPa	
Linear	436	2200
0.2% offset Yield strain	0.02116 mm/mm	
0.2% offset Yield stress	114.64 MPa	
Toughness	3.80 MJ/mm ³	

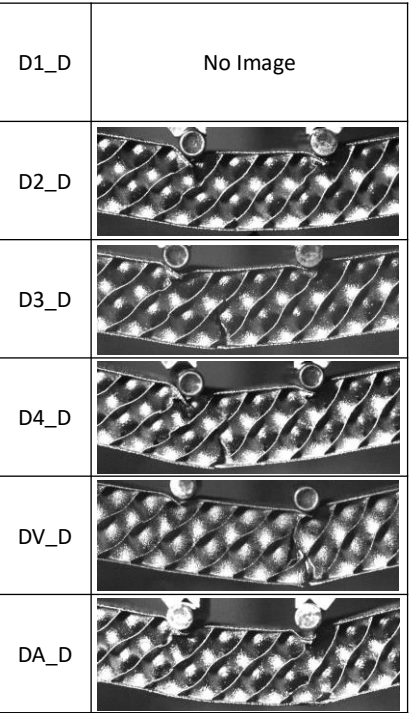
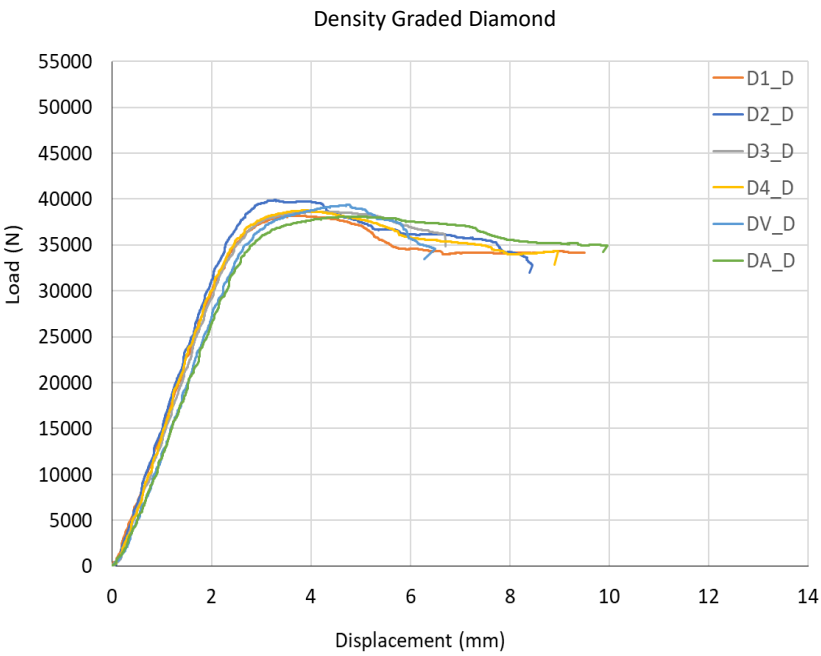
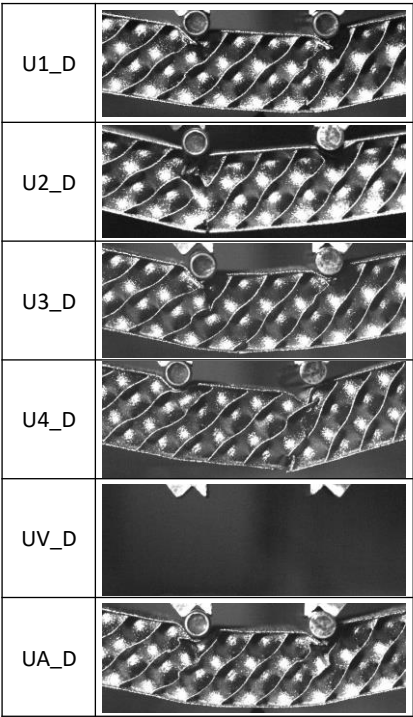
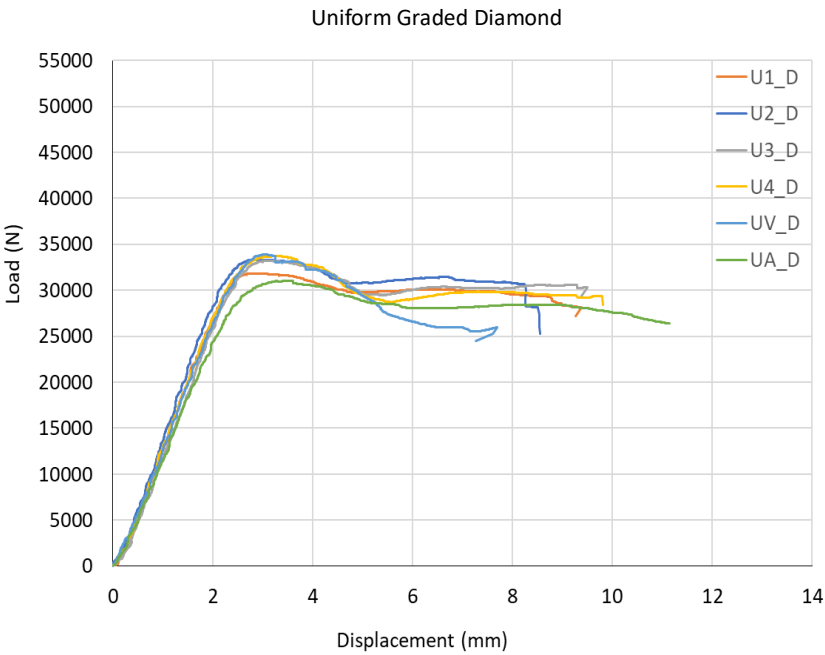


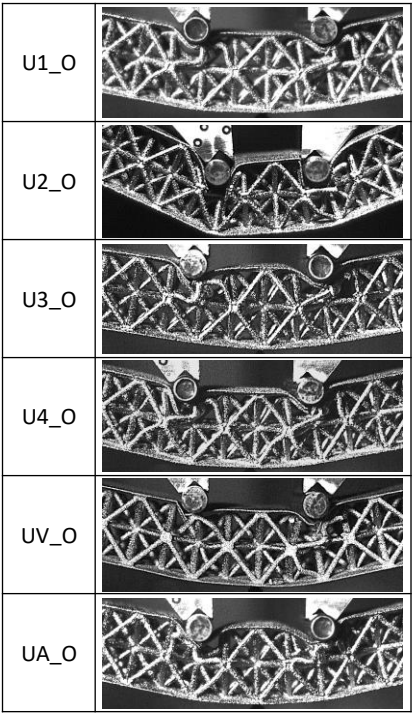
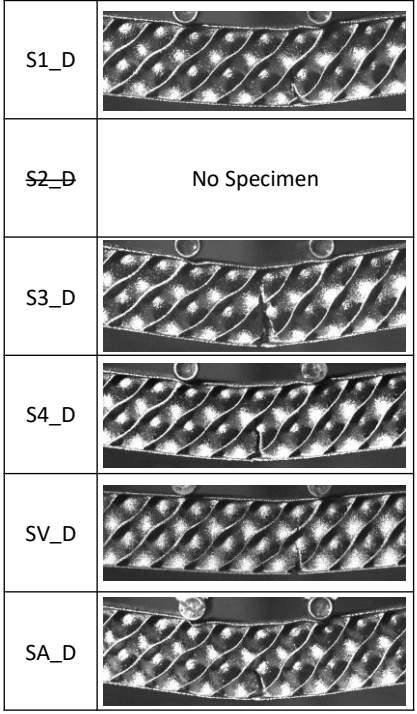
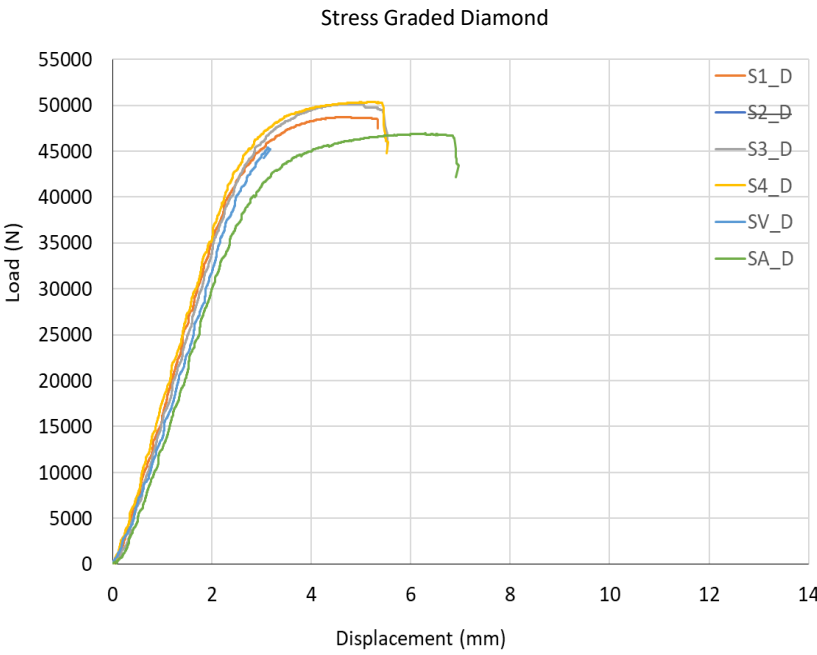
Load displacement curves and resulting images

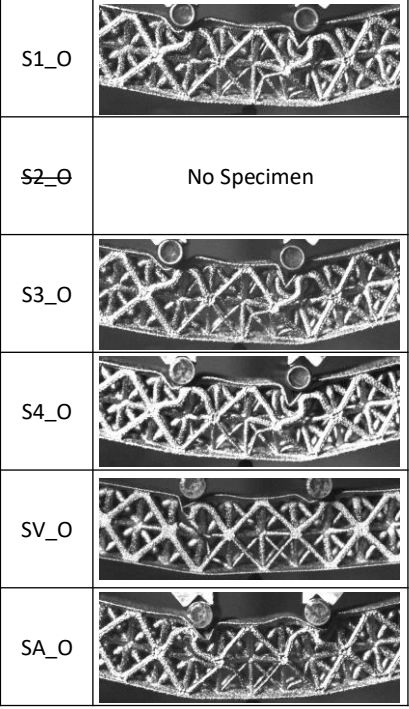
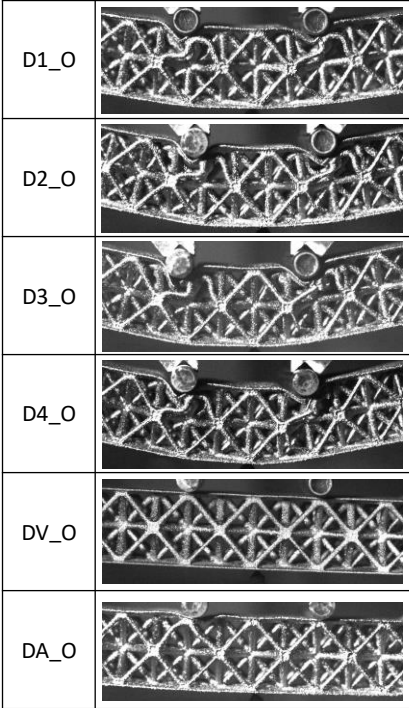
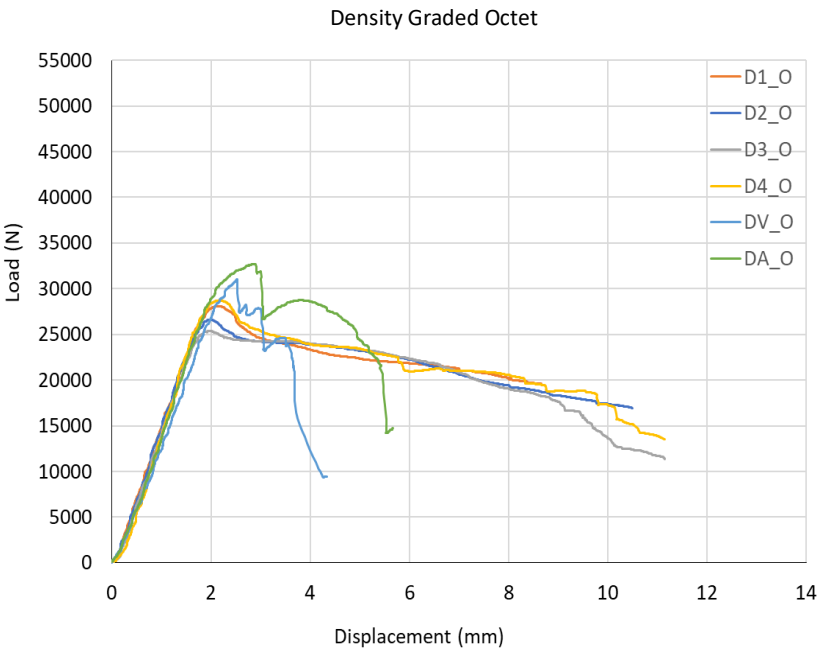


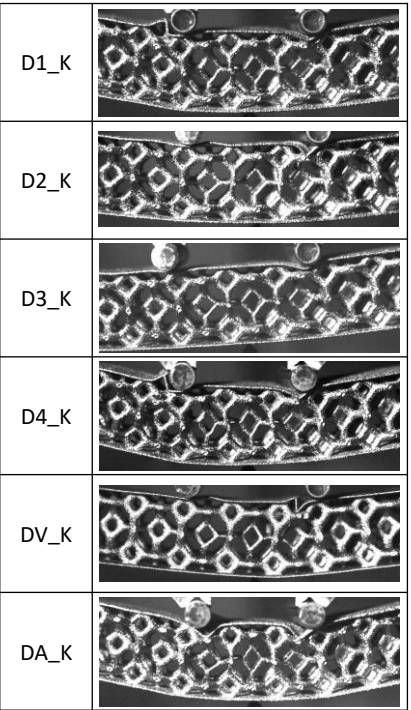
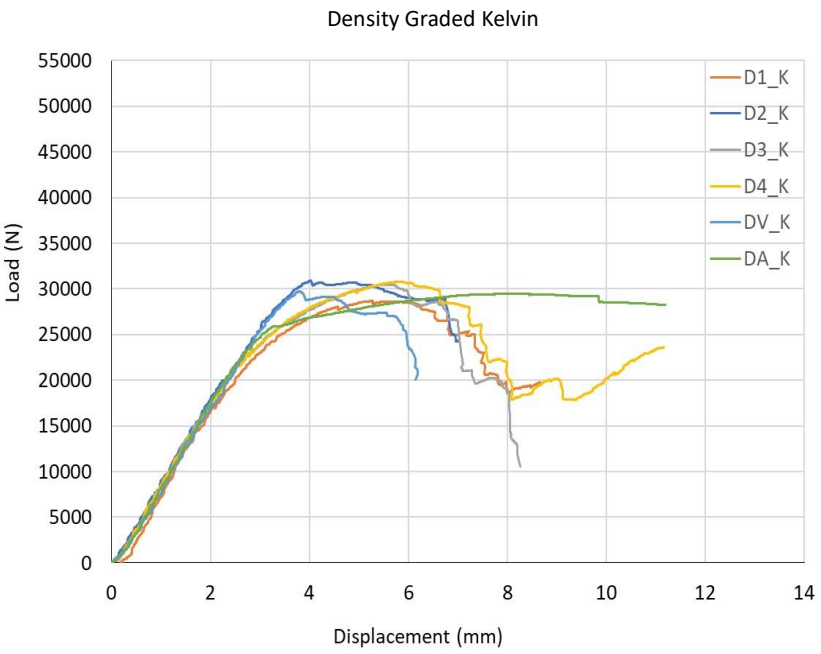
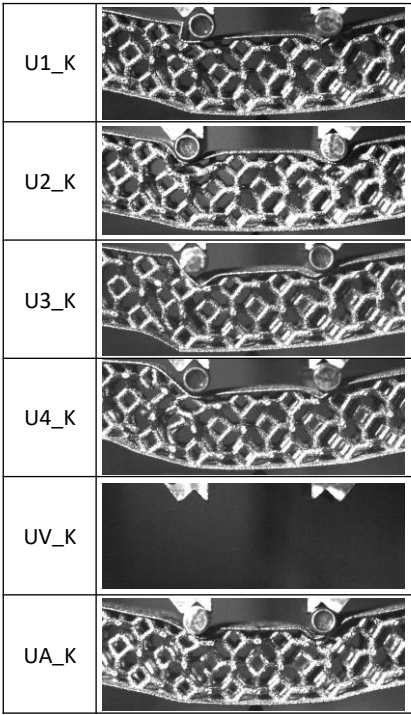
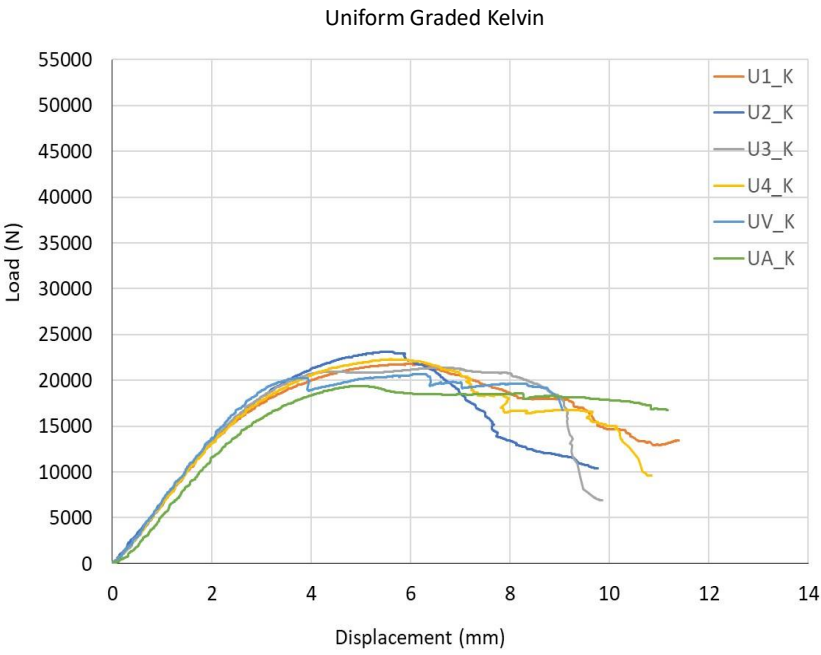


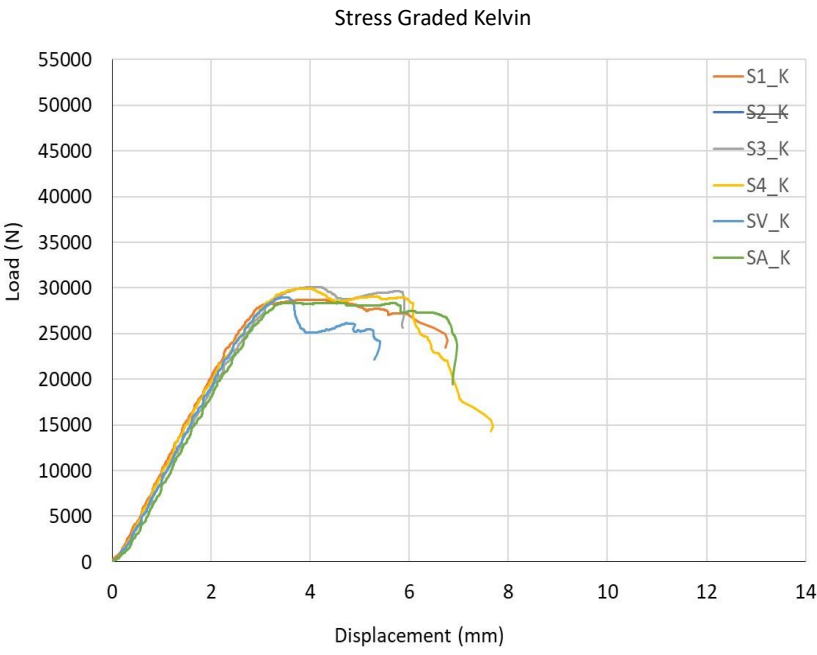






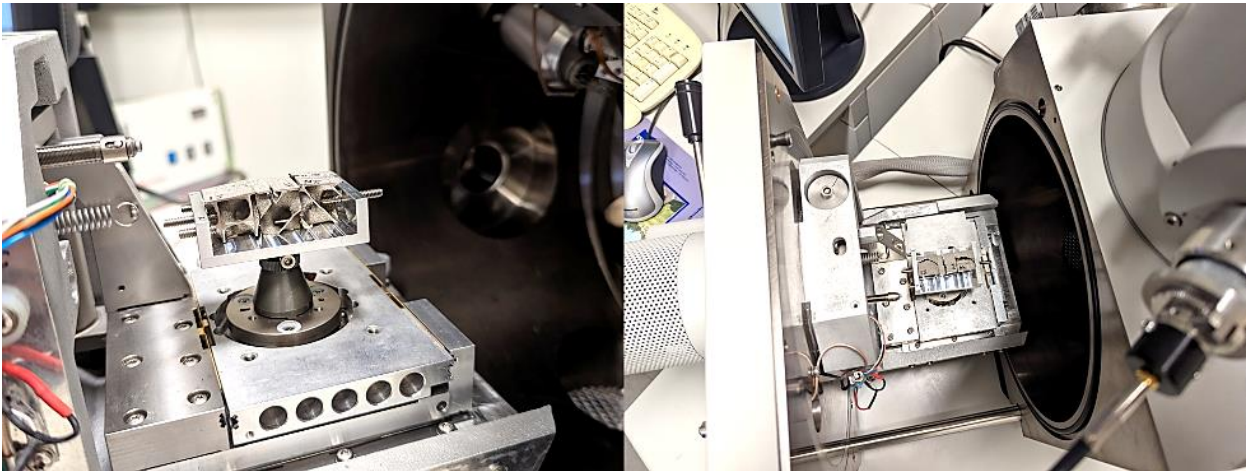


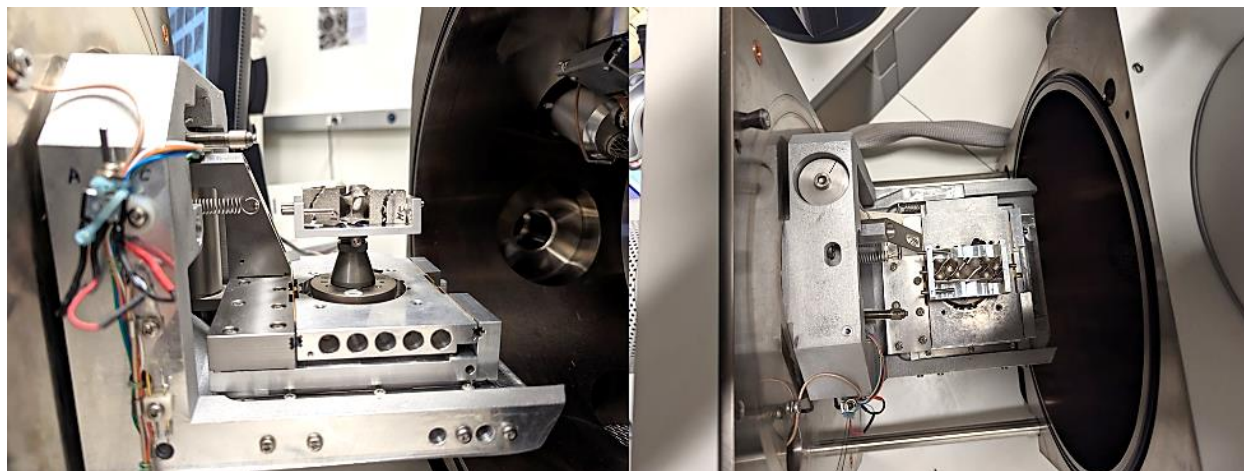




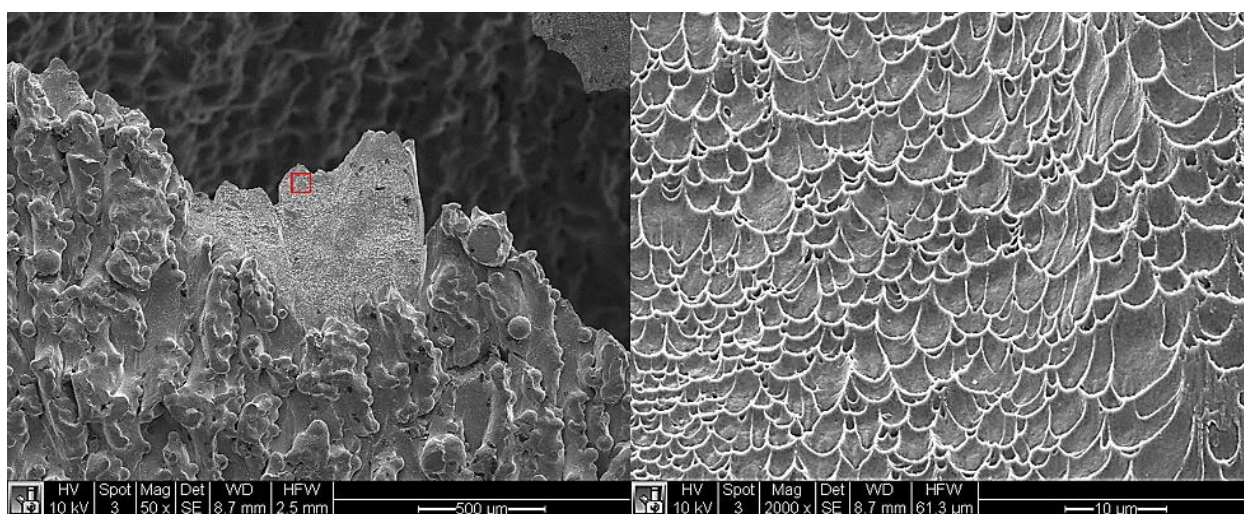
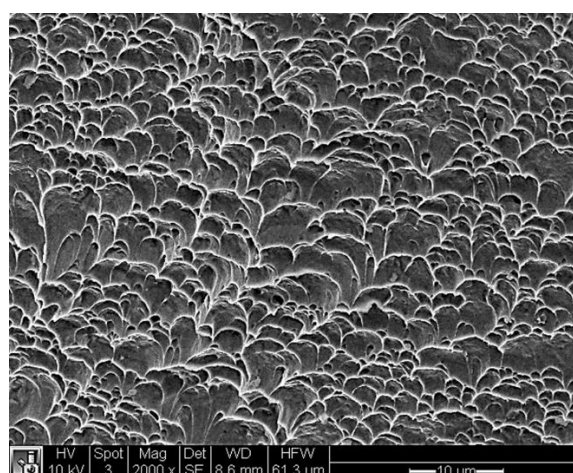
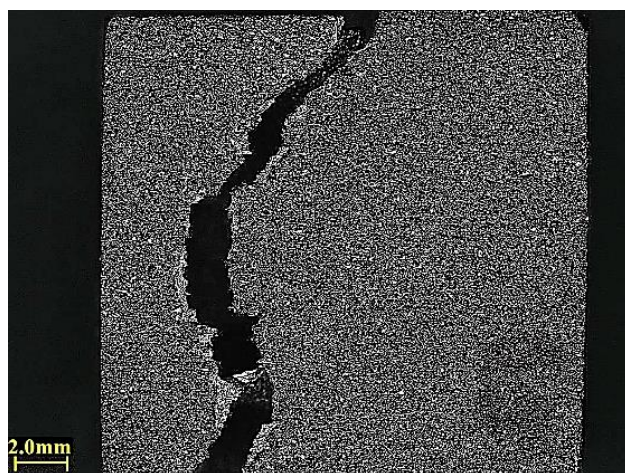
S1_K	
S2_K	No Specimen
S3_K	
S4_K	
SV_K	
SA_K	

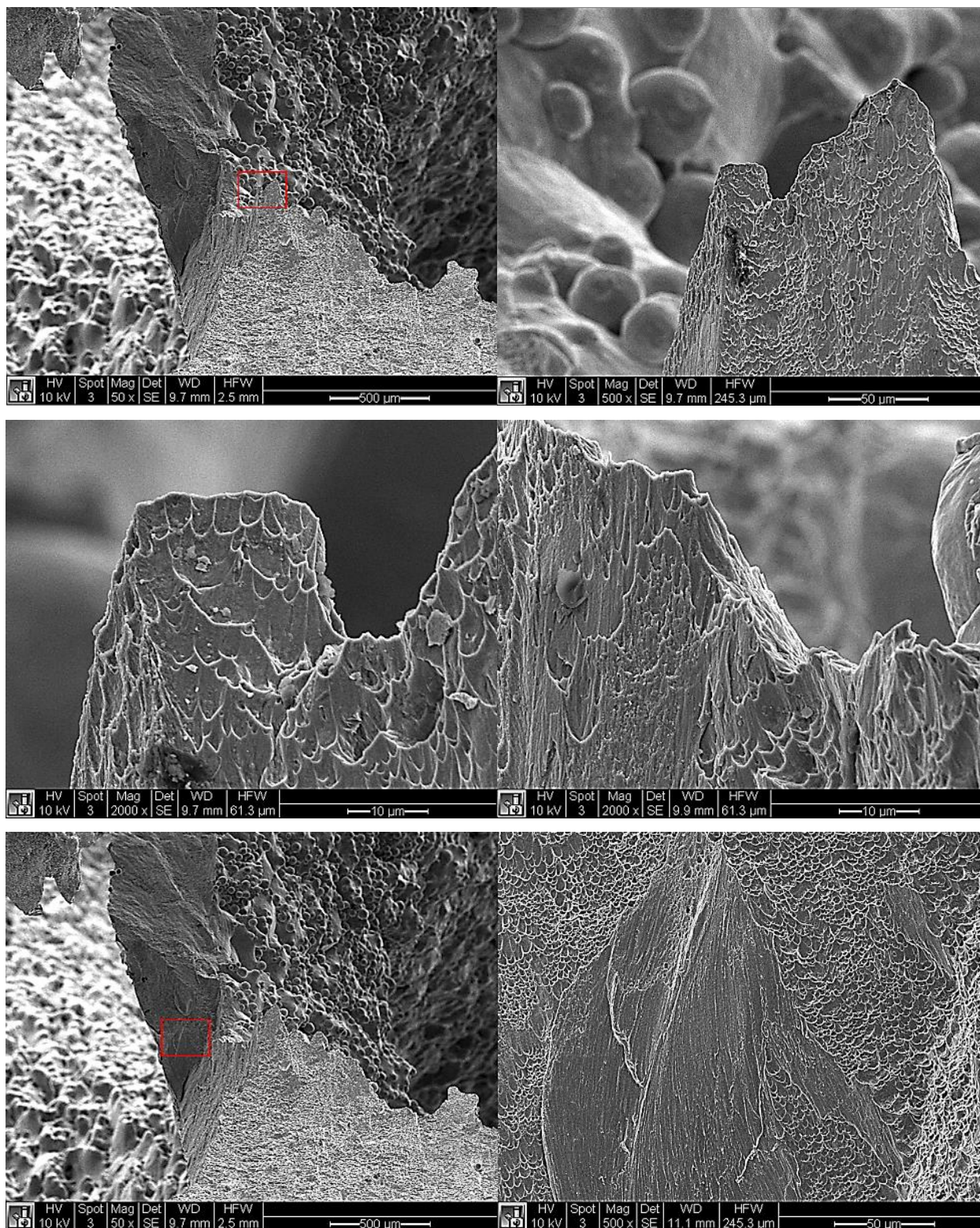
SEM fracture surface inspection setup

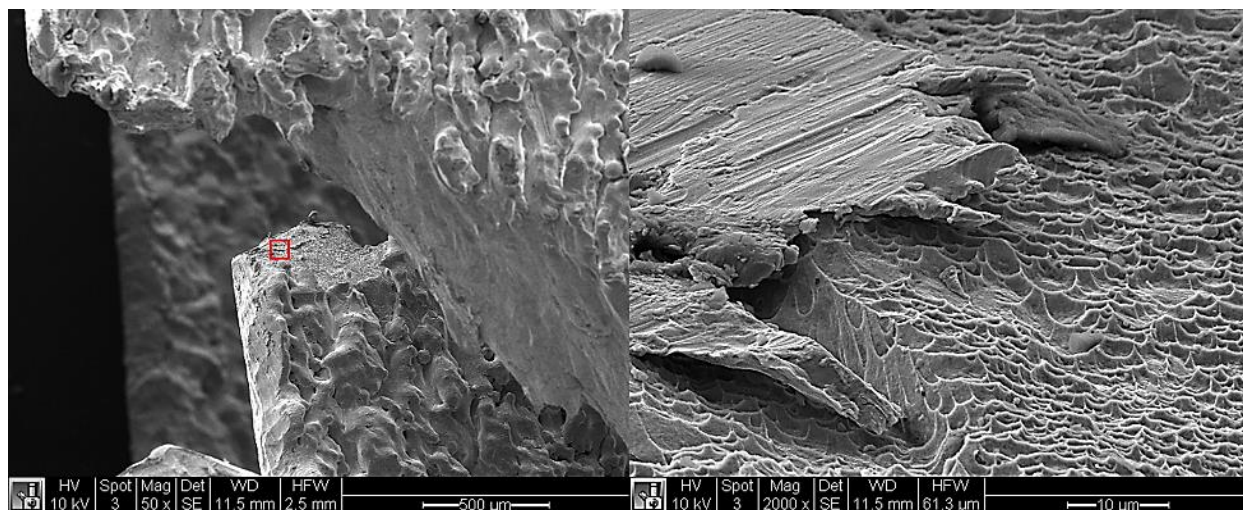




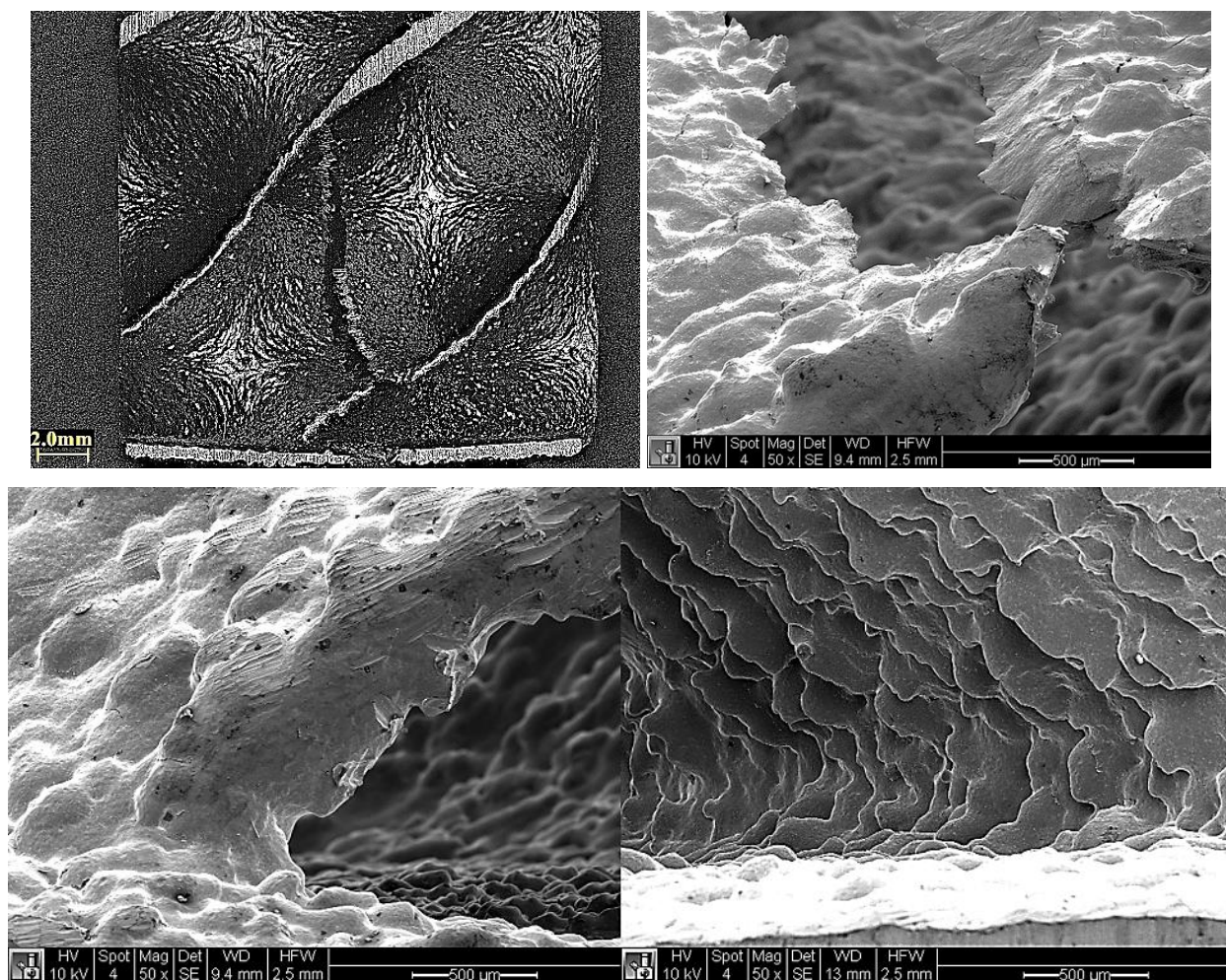
Specimen S1_D (Stress graded Diamond from Build 1) bottom view

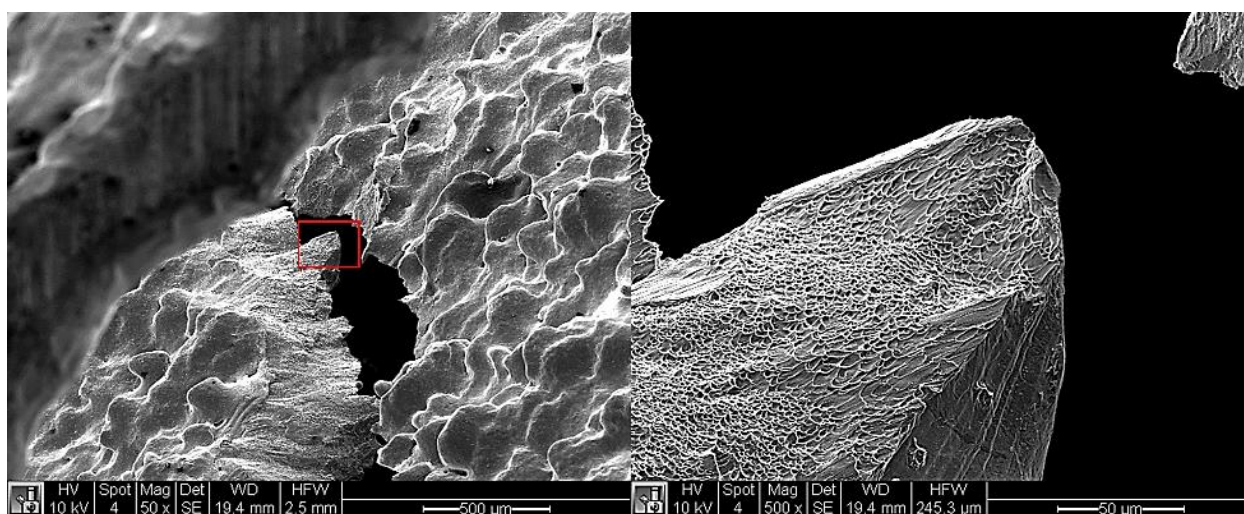
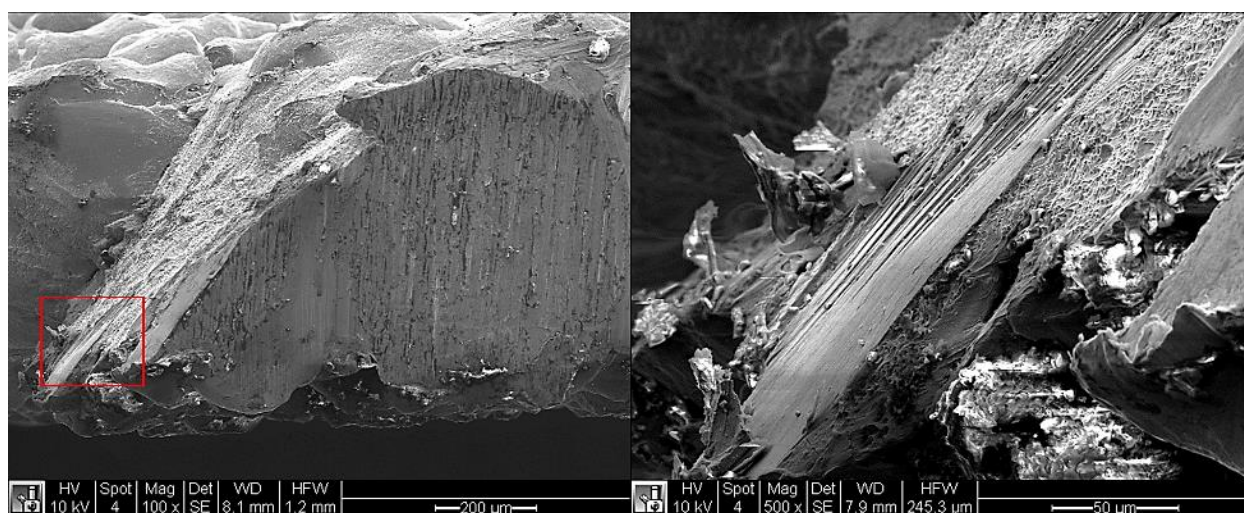
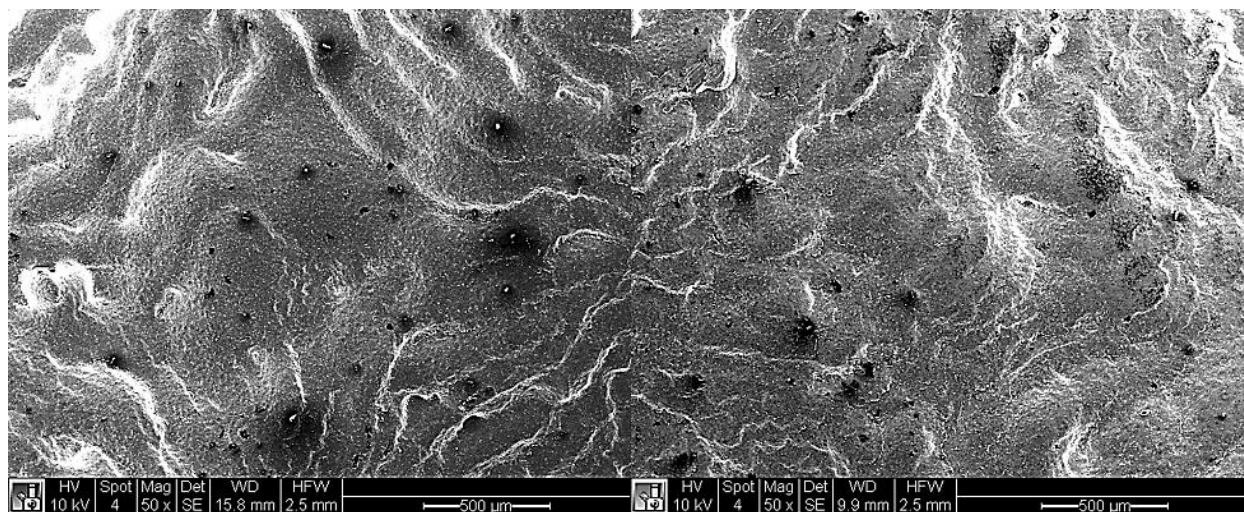


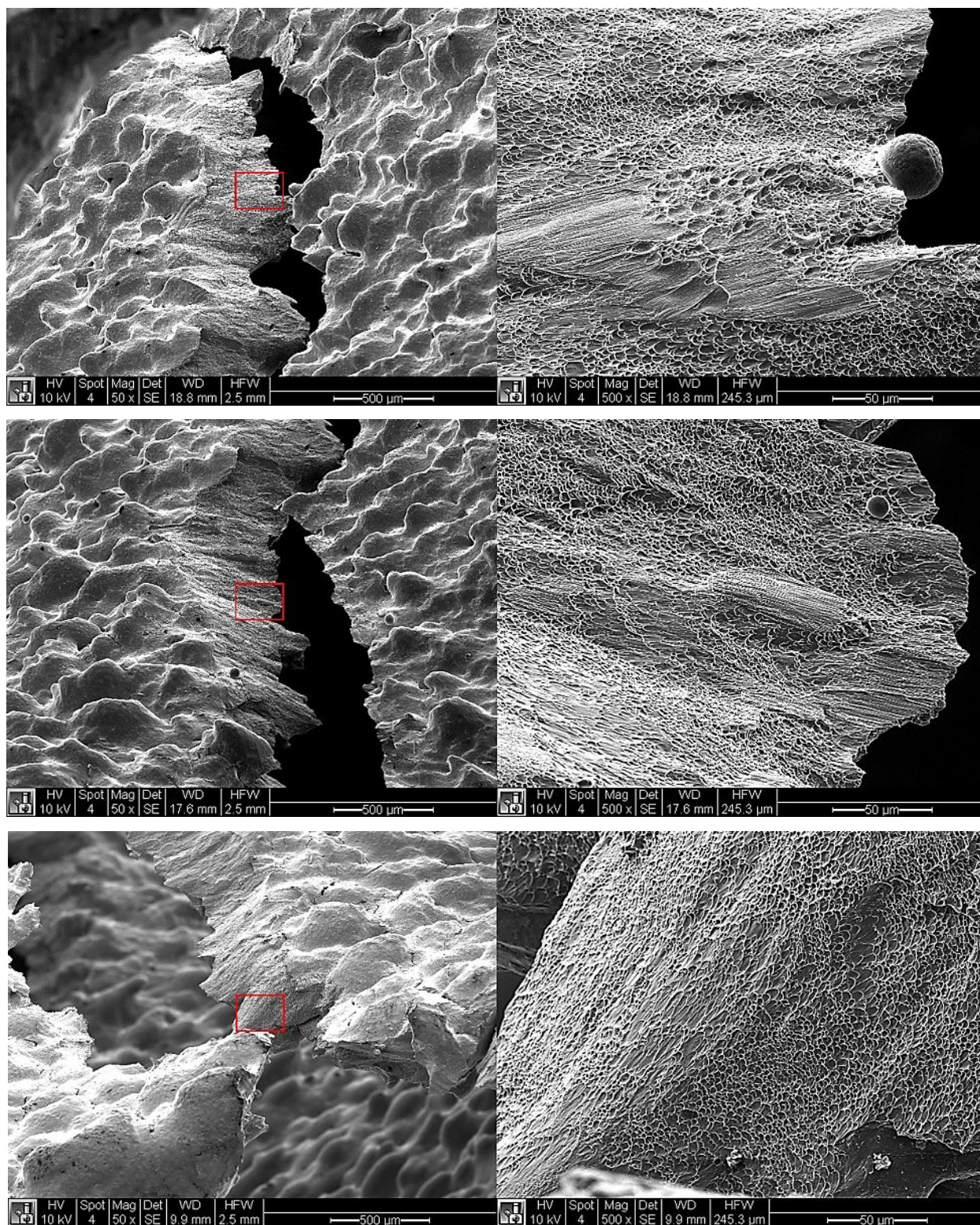


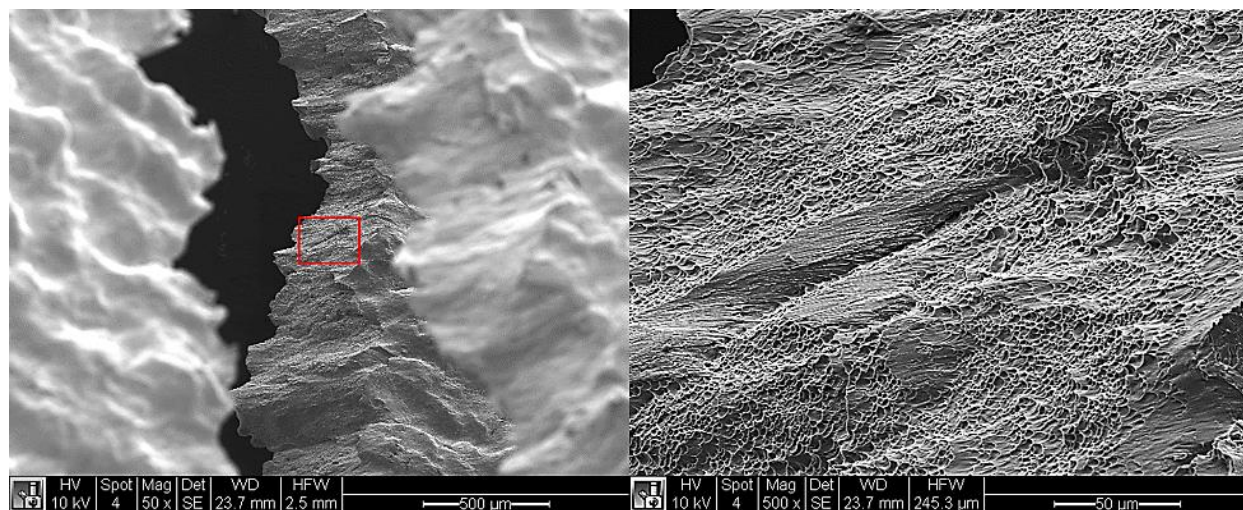


Specimen S1_D (Stress graded Diamond from Build 1) side view

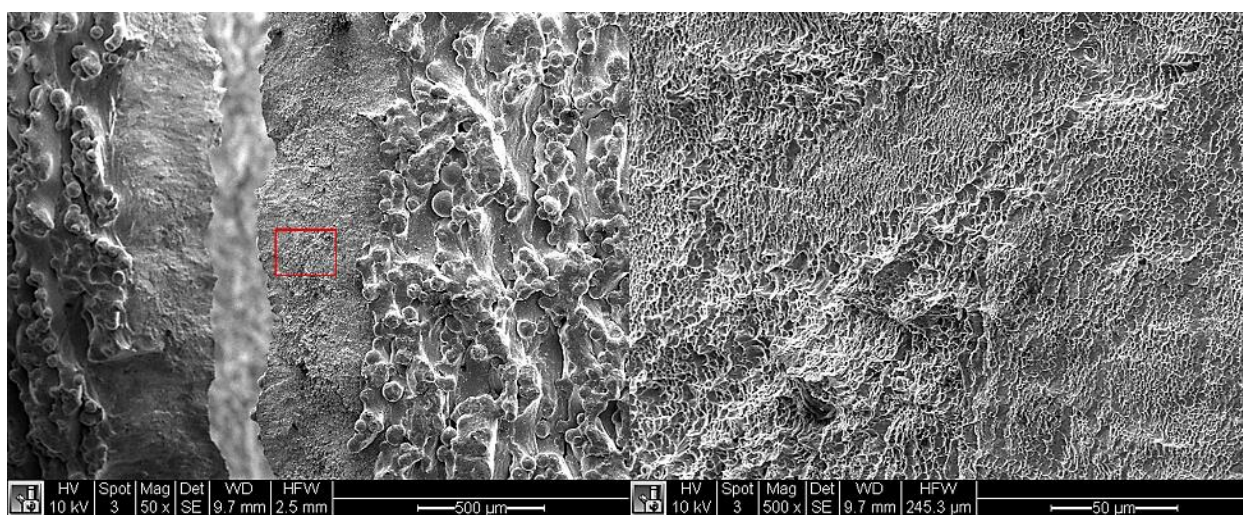
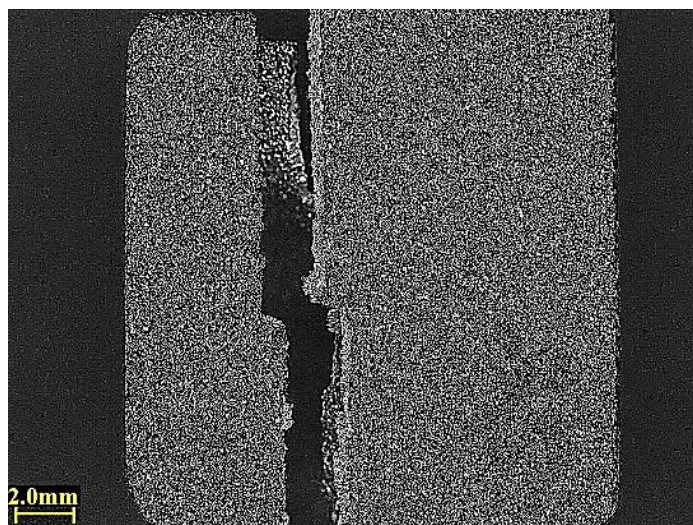


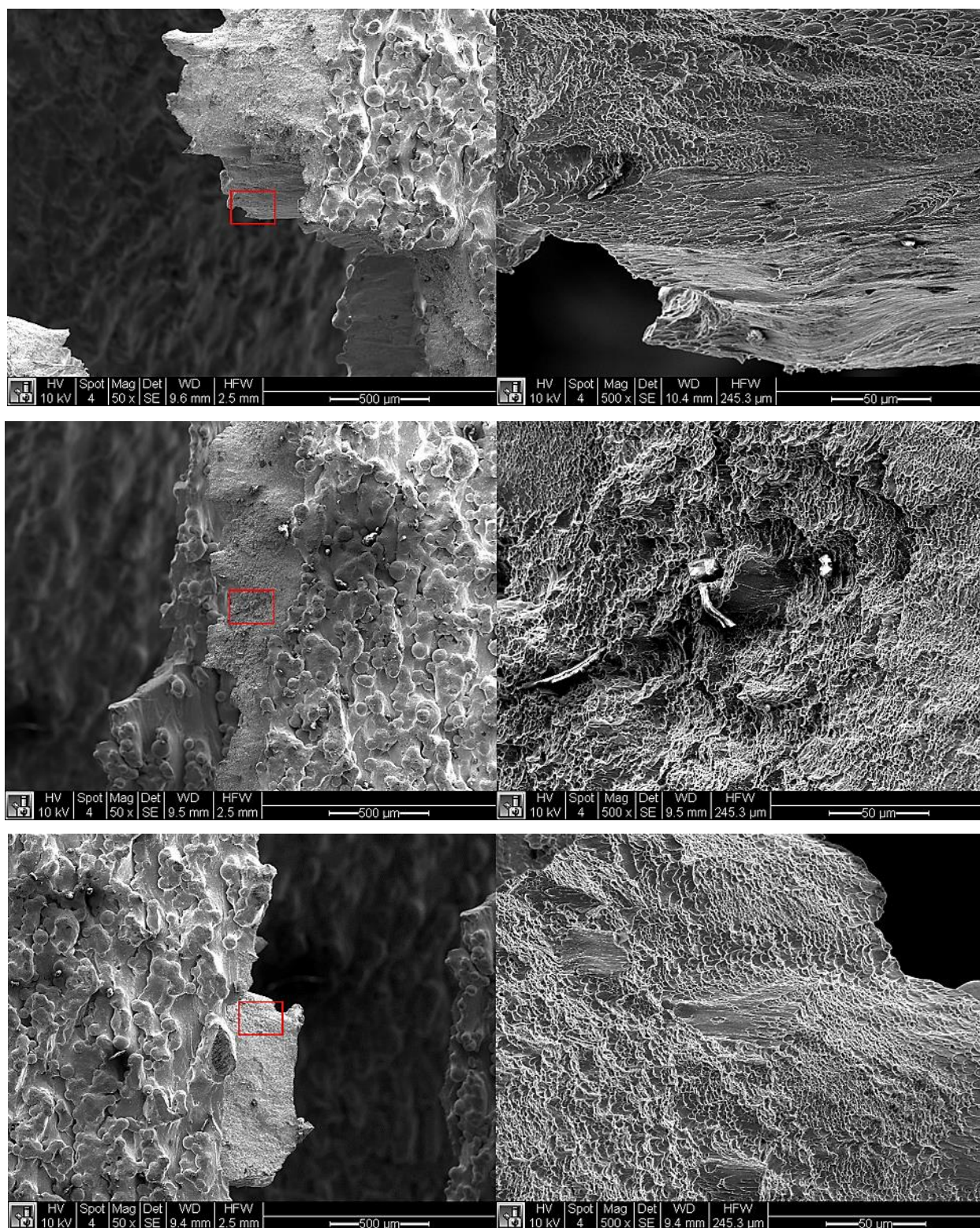


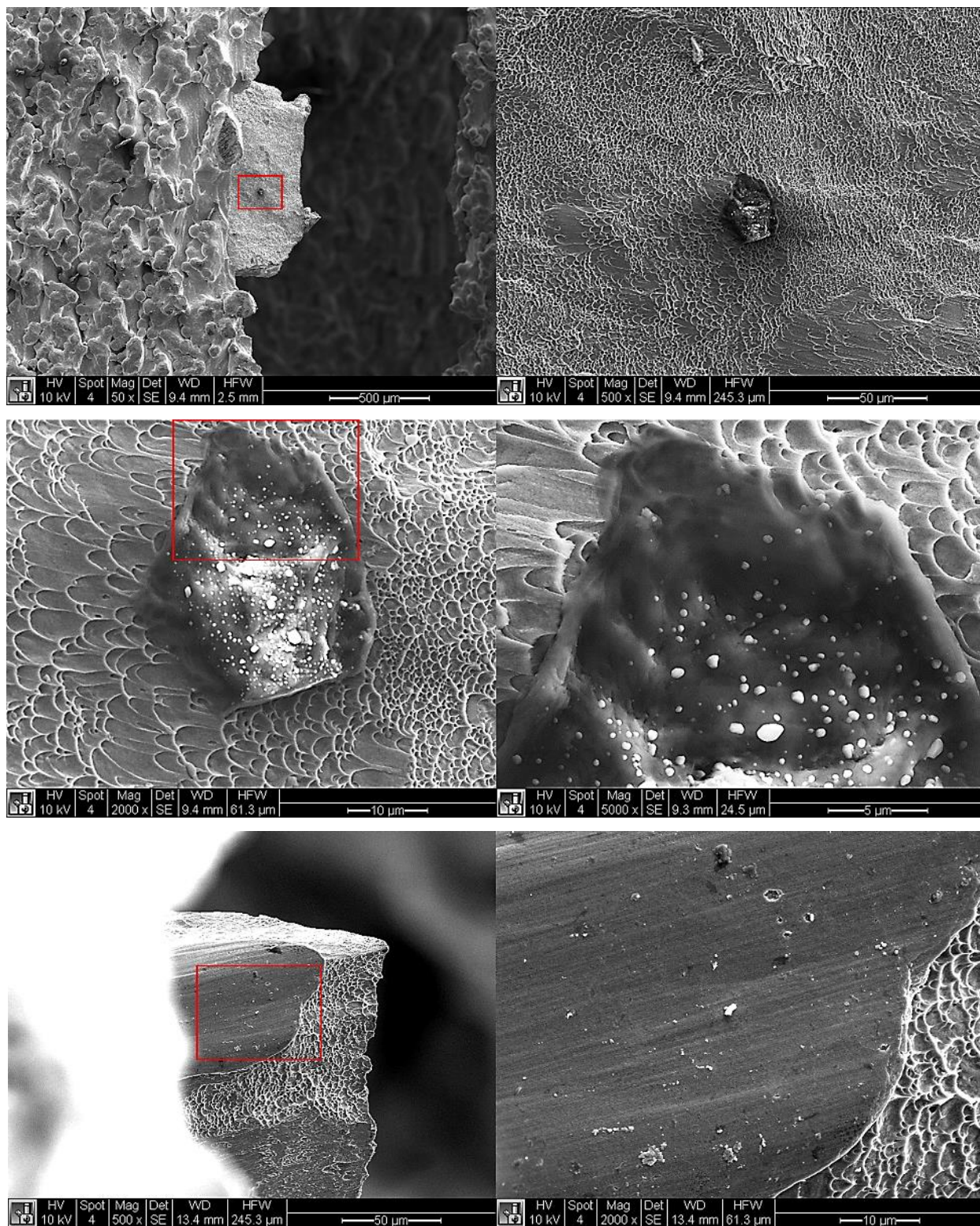




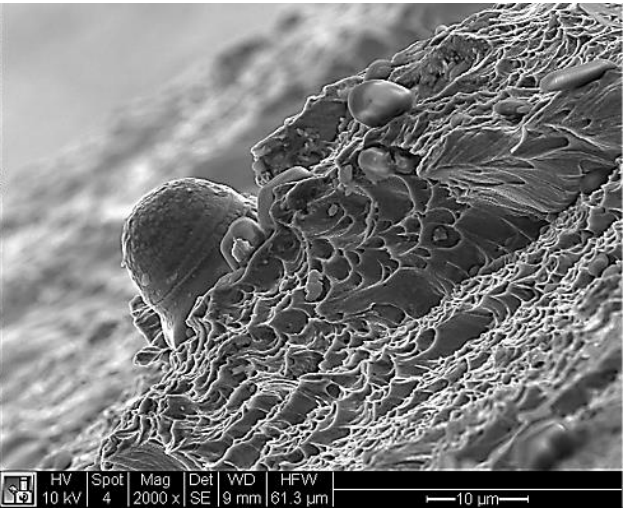
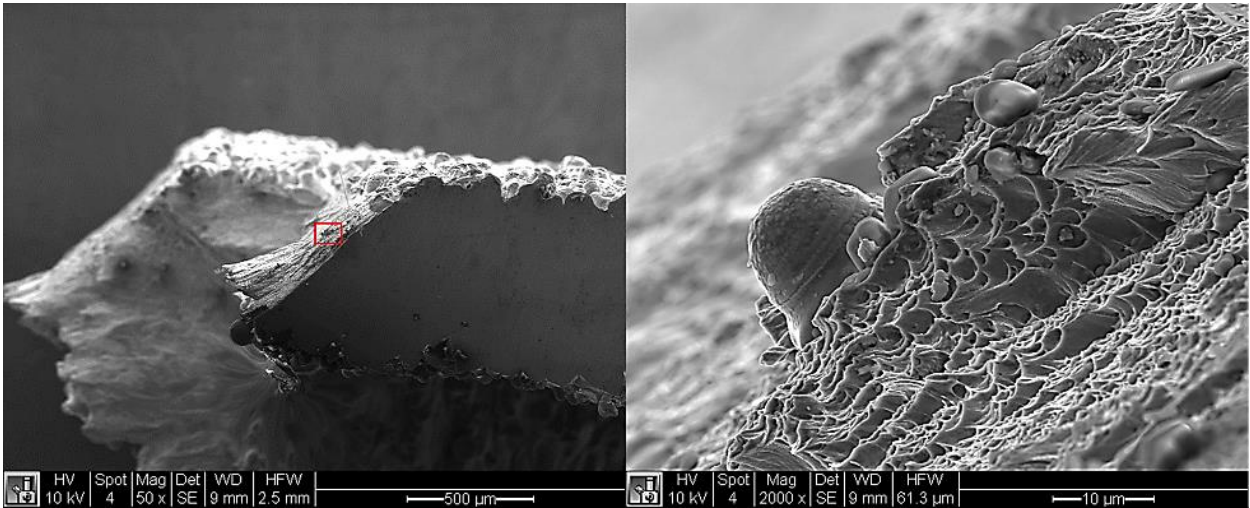
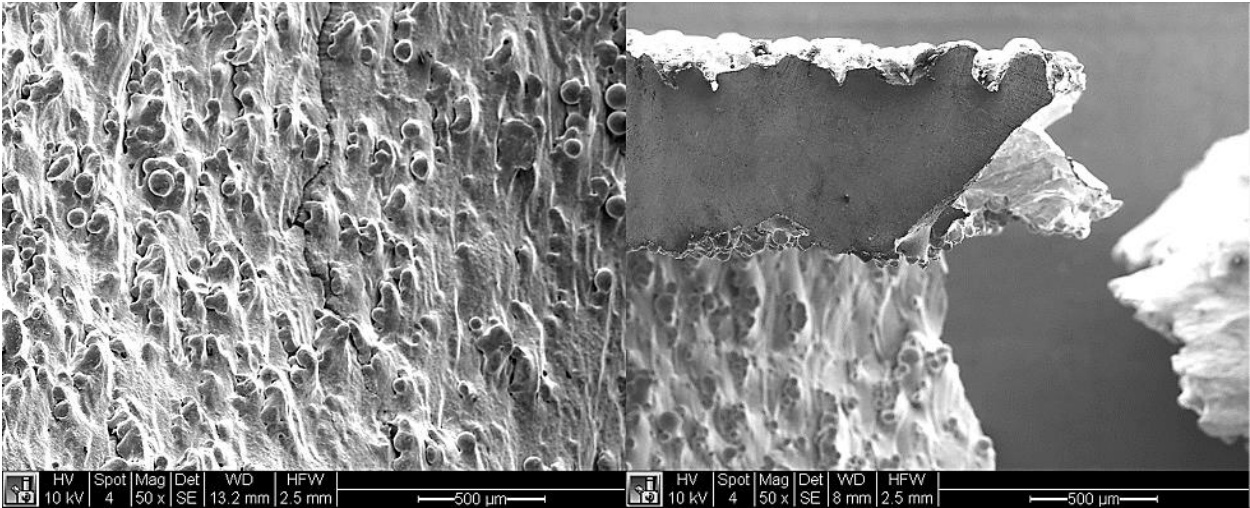
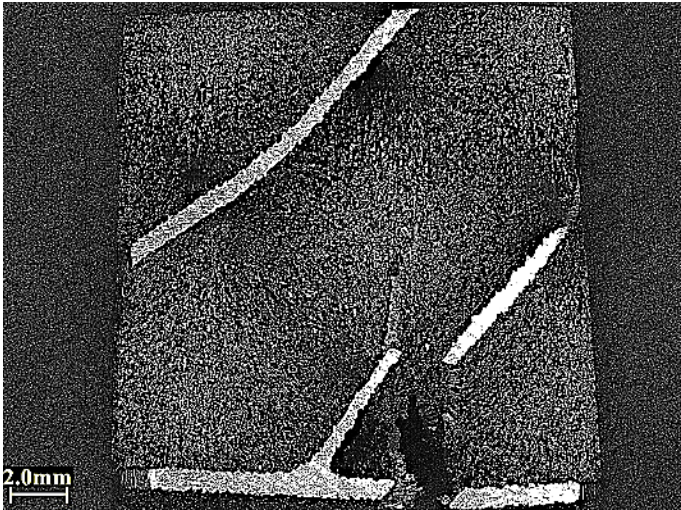
Specimen SV_D (Stress graded Diamond from Build Vertical) bottom view

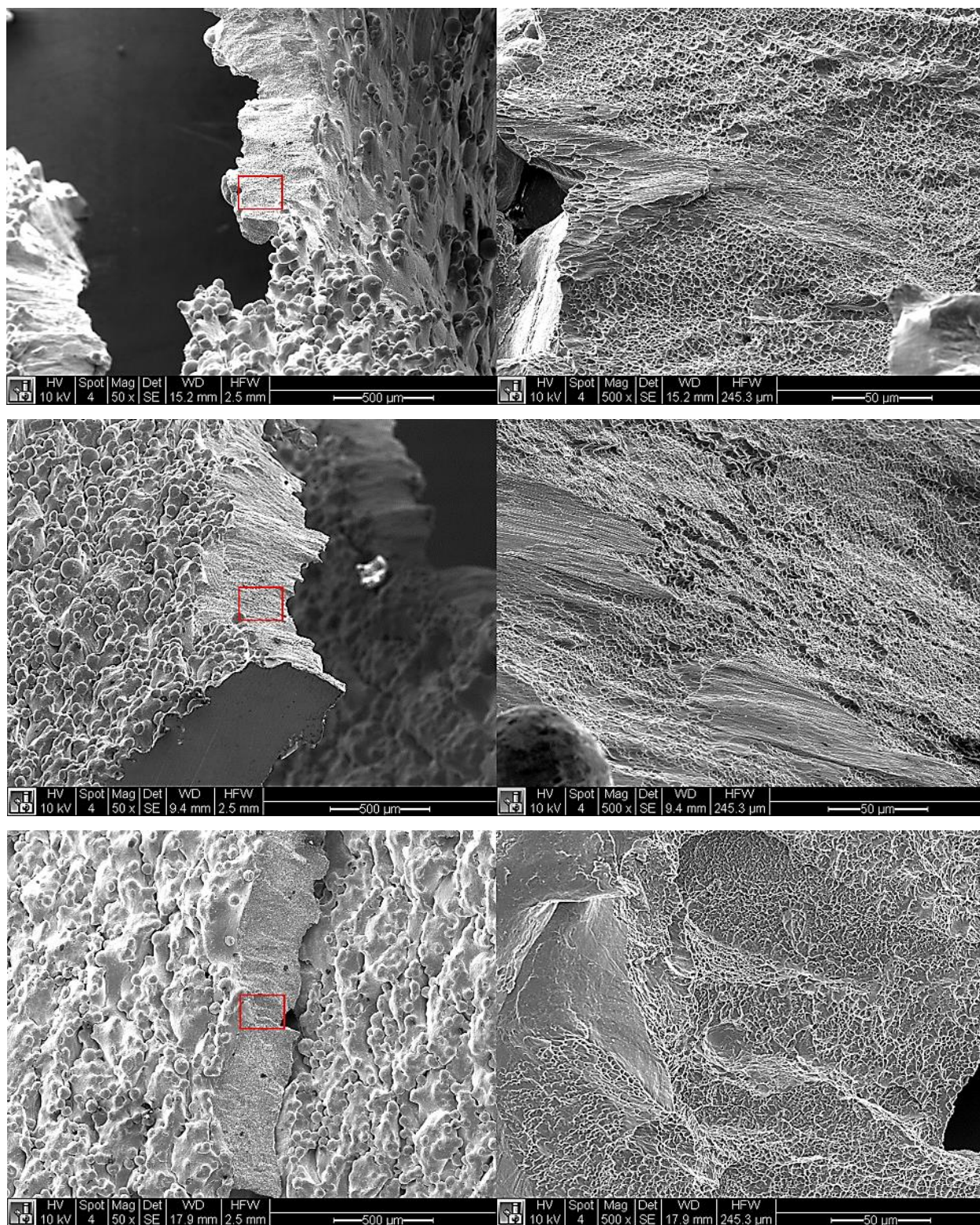


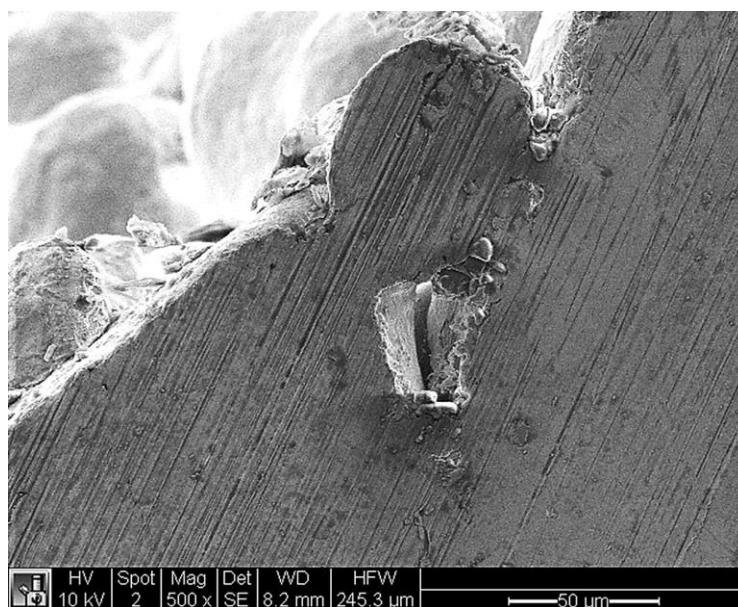
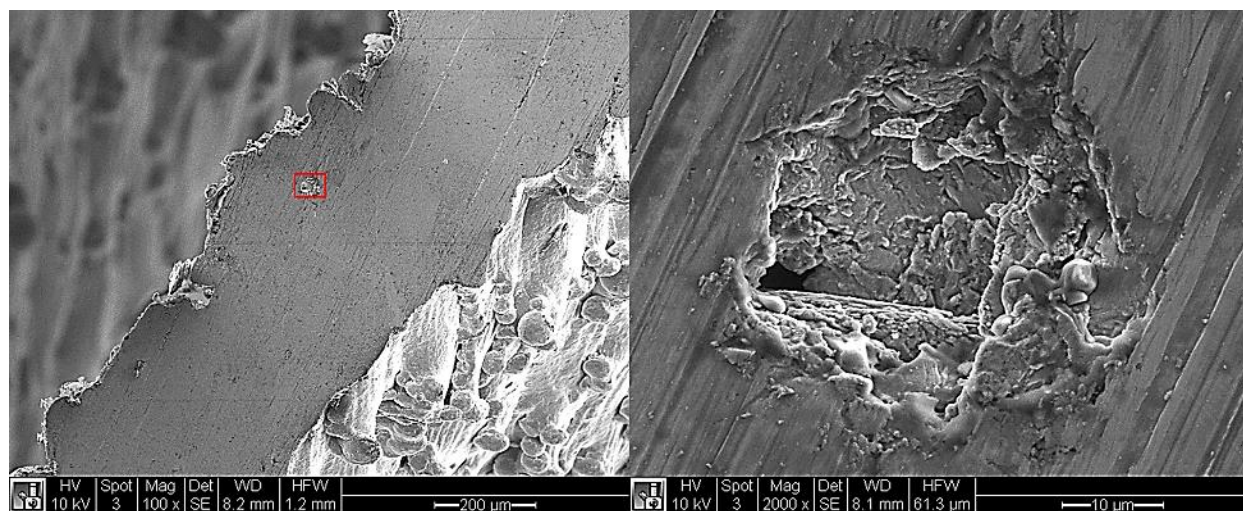
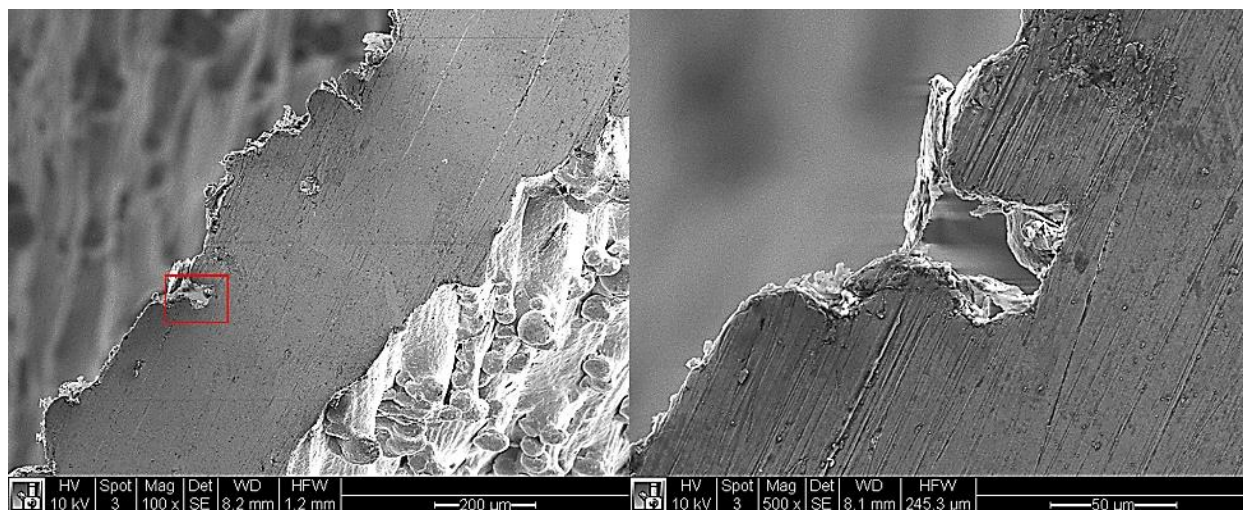




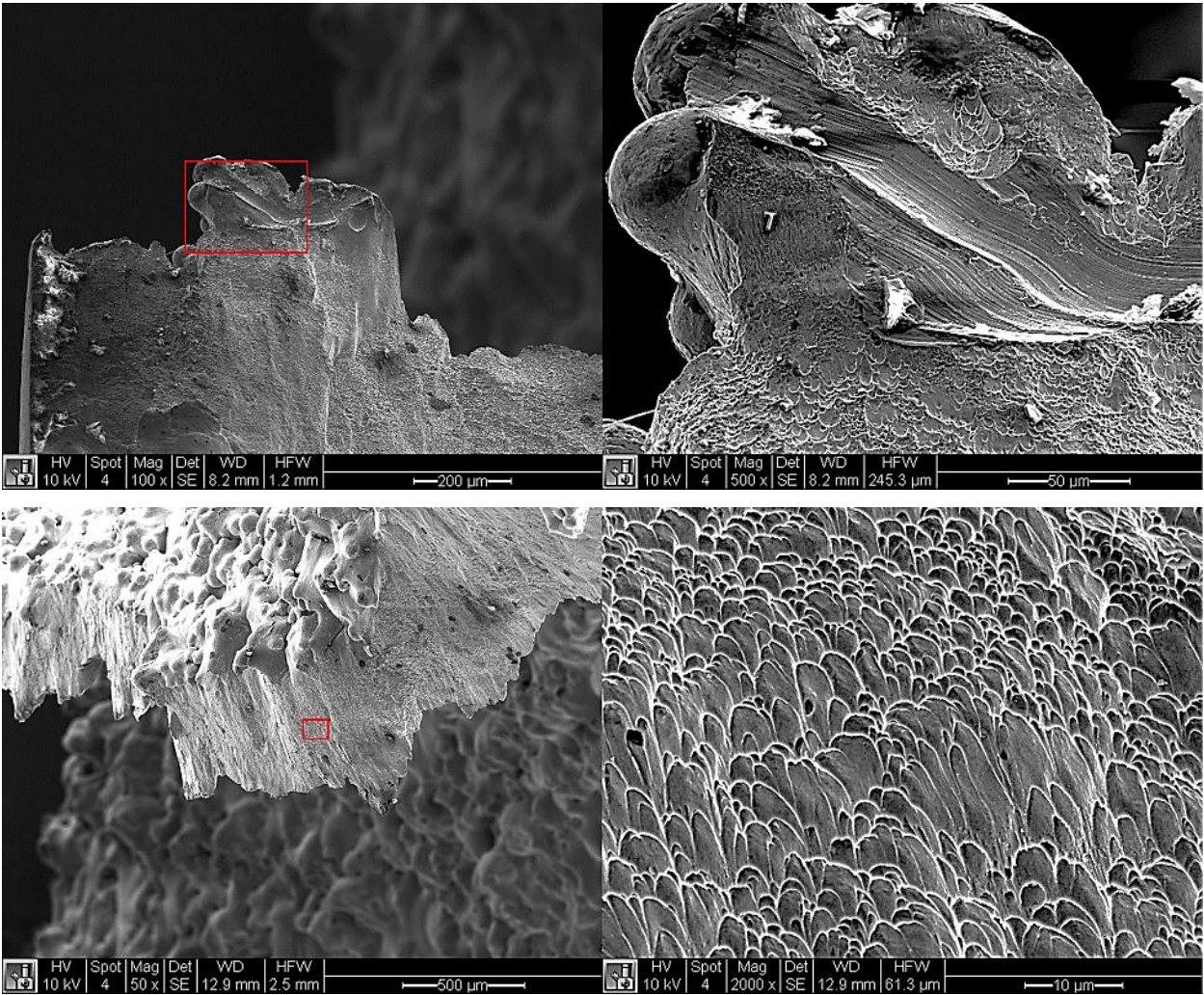
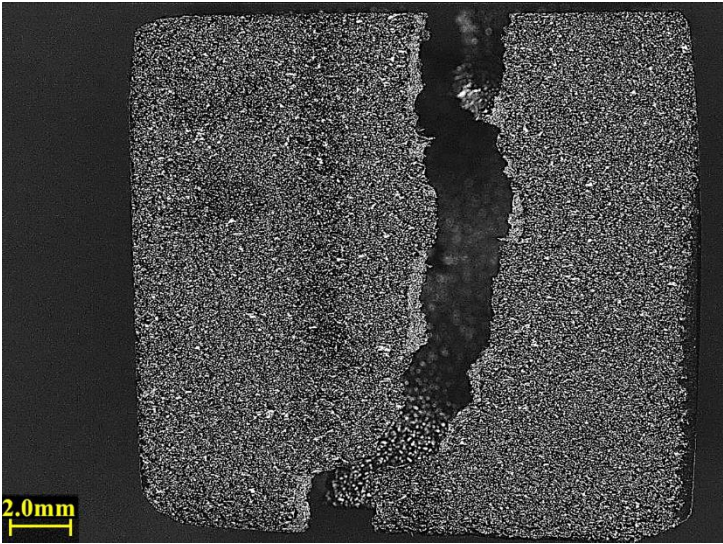
Specimen SV_D (Stress graded Diamond from Build Vertical) side view

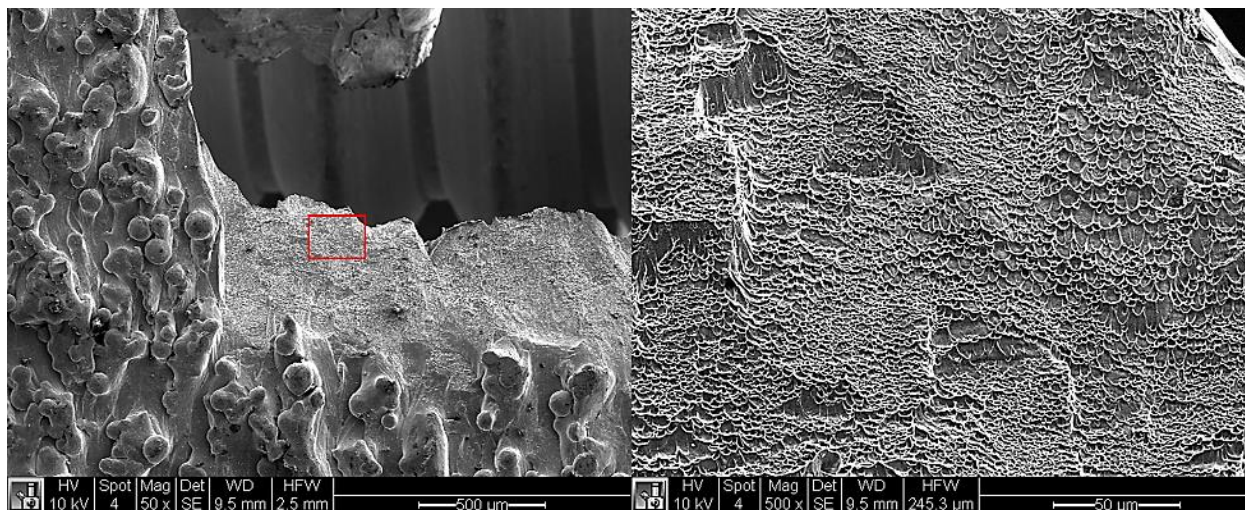




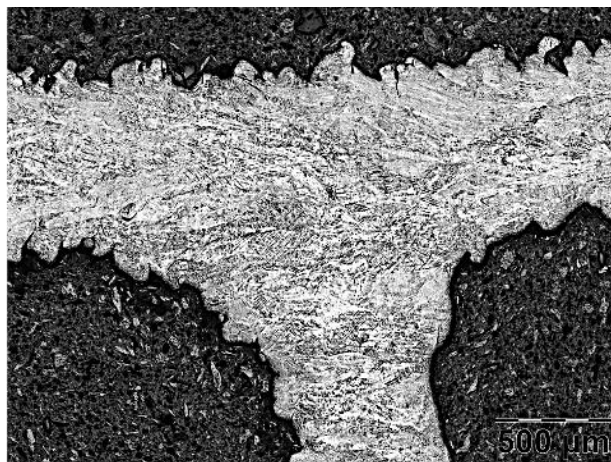
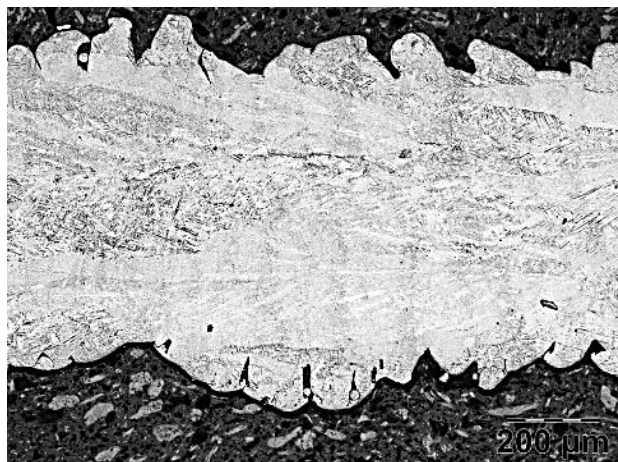


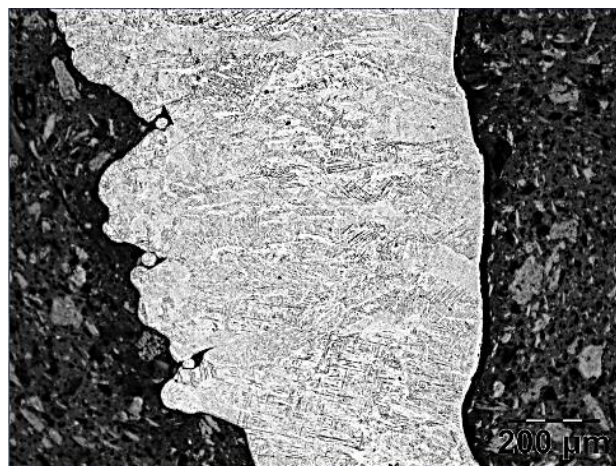
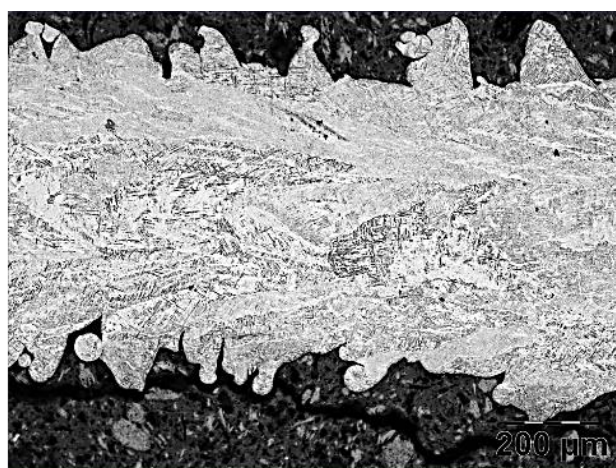
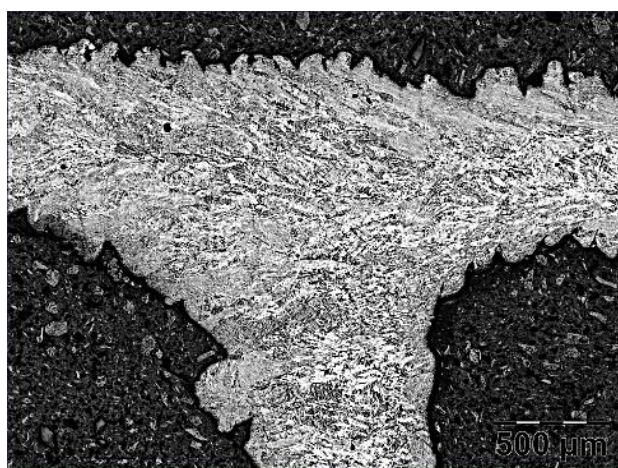
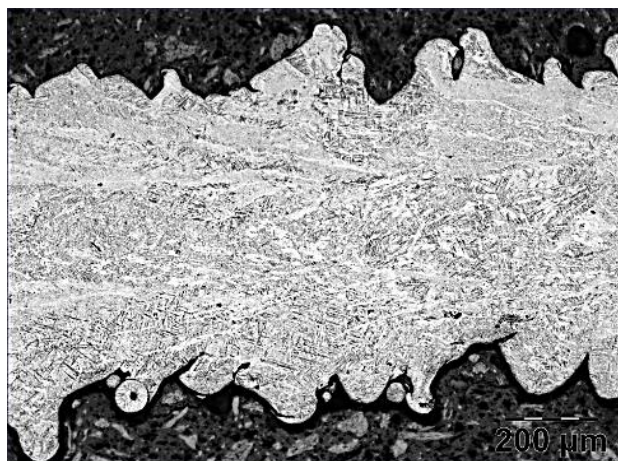
Specimen SA_D (Stress graded Diamond from Build Arcam) bottom view

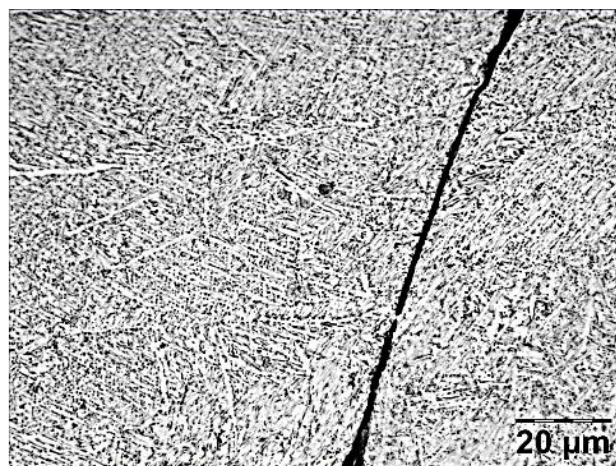
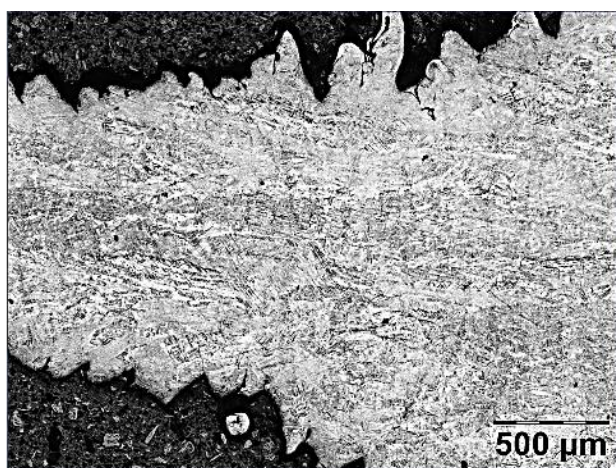
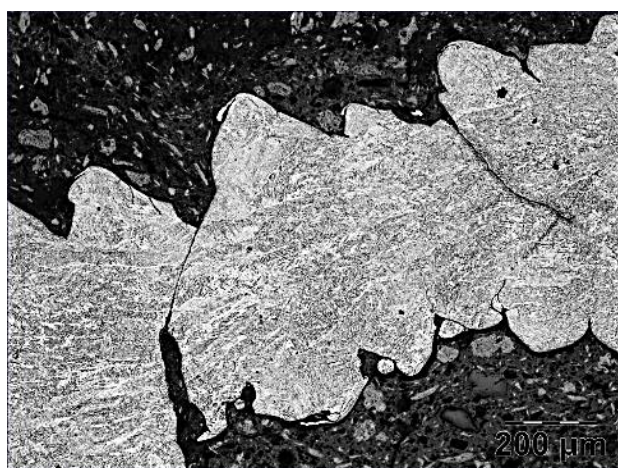
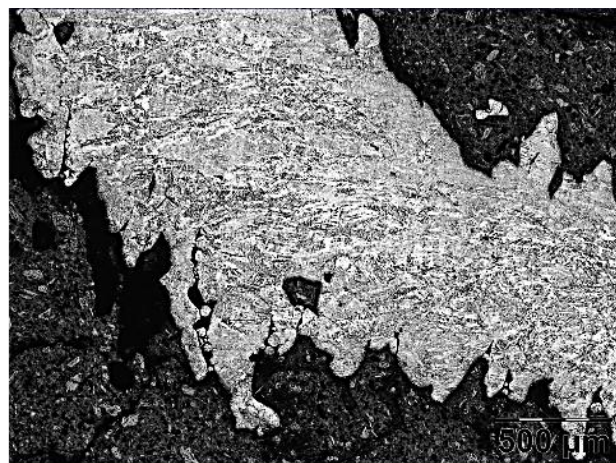


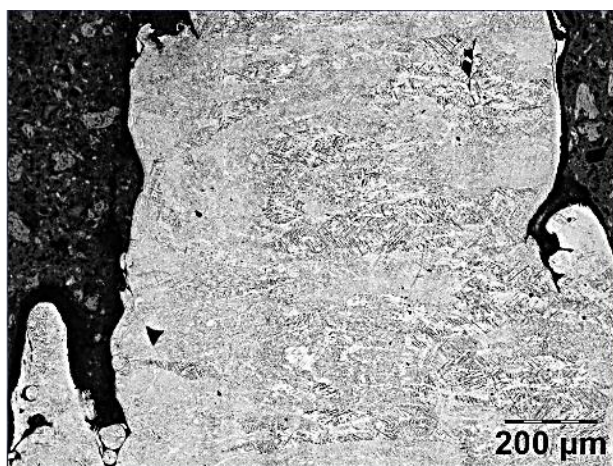
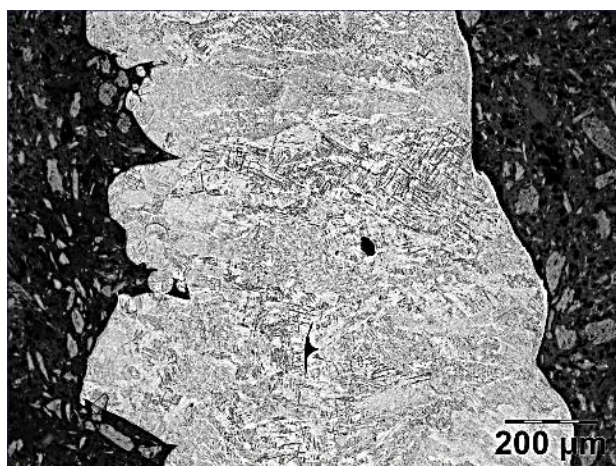
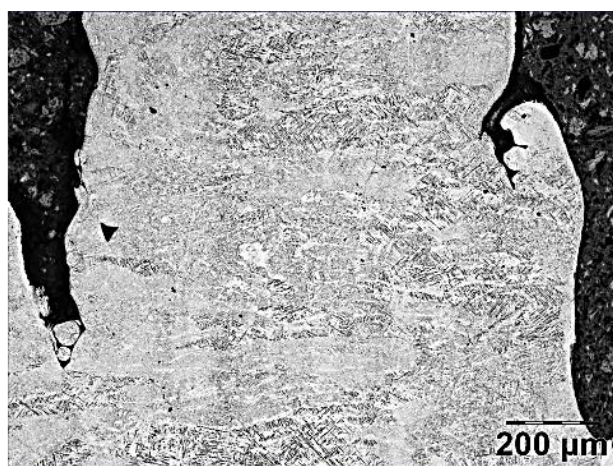
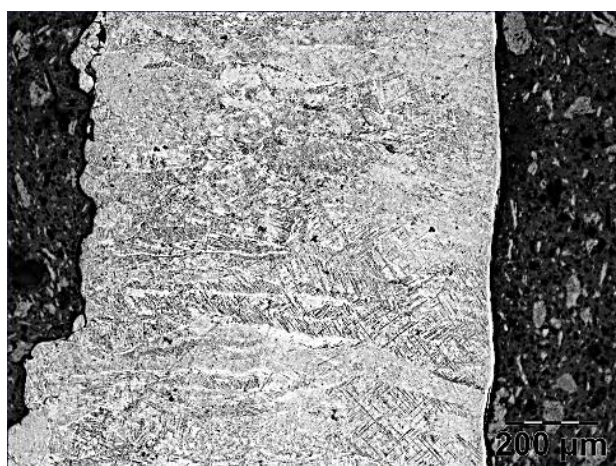
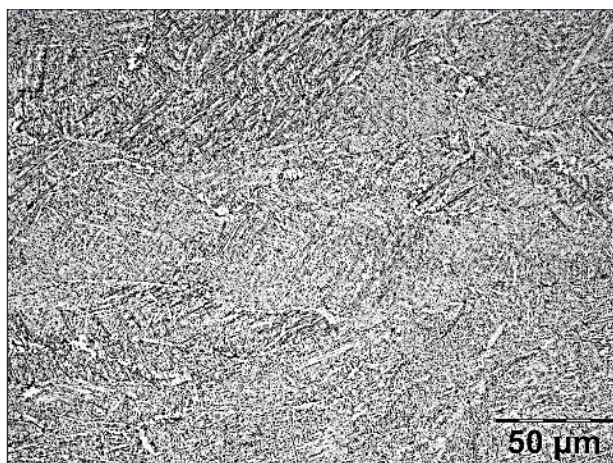
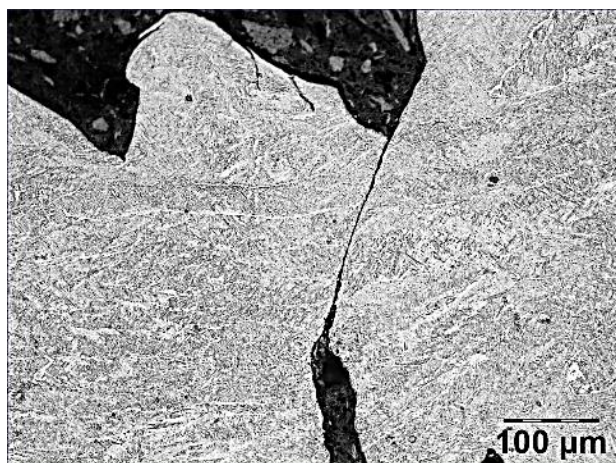


Specimen S1_D (Stress graded Diamond from Build 1) Build Direction Left

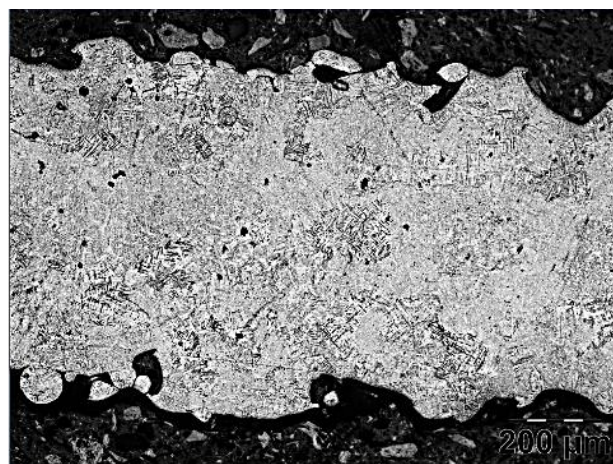
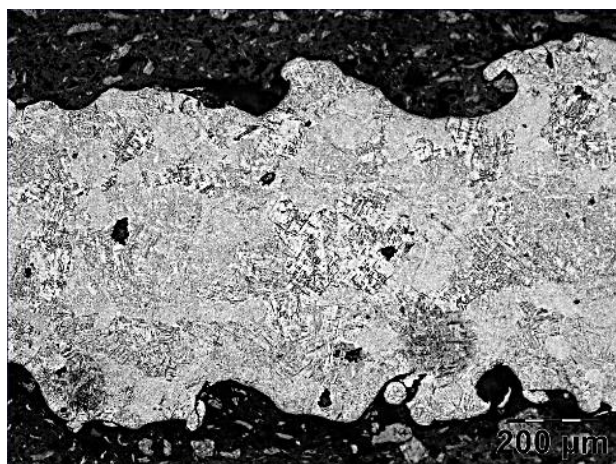
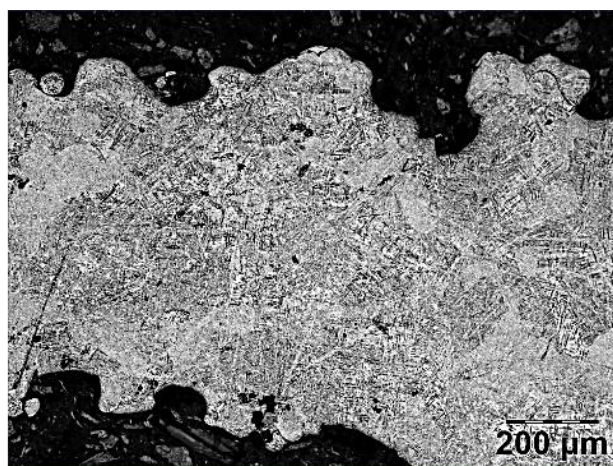


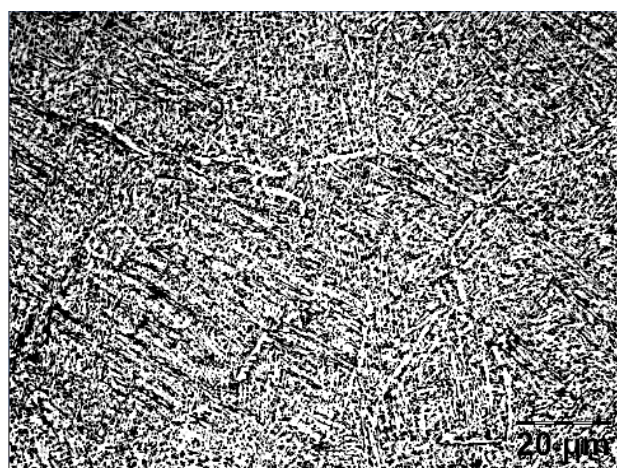
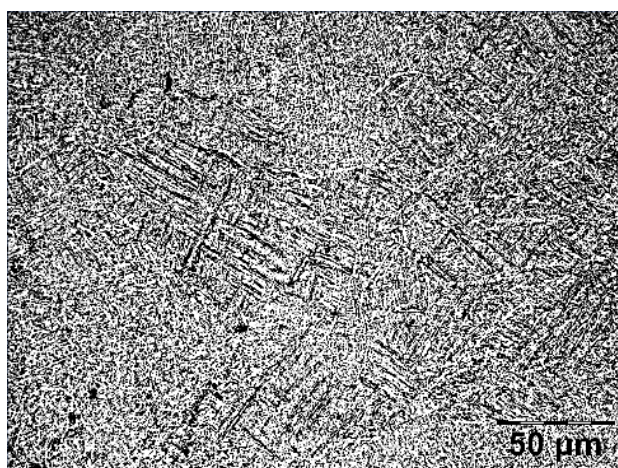
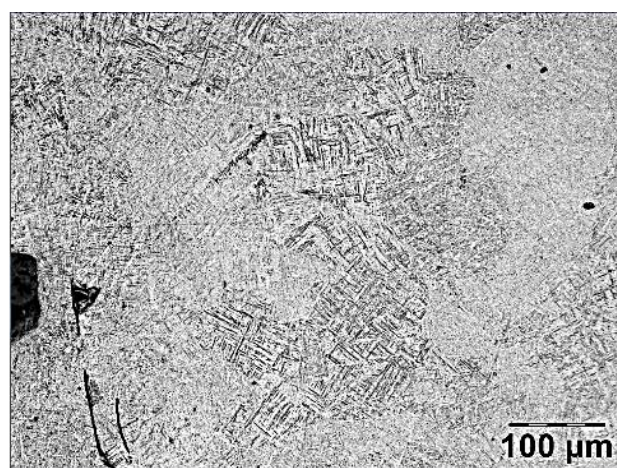
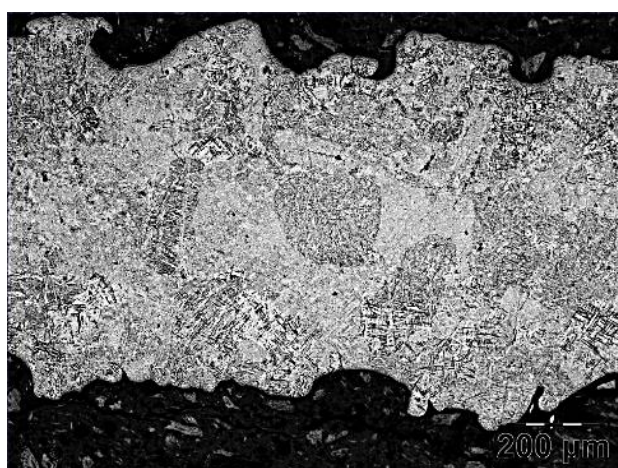
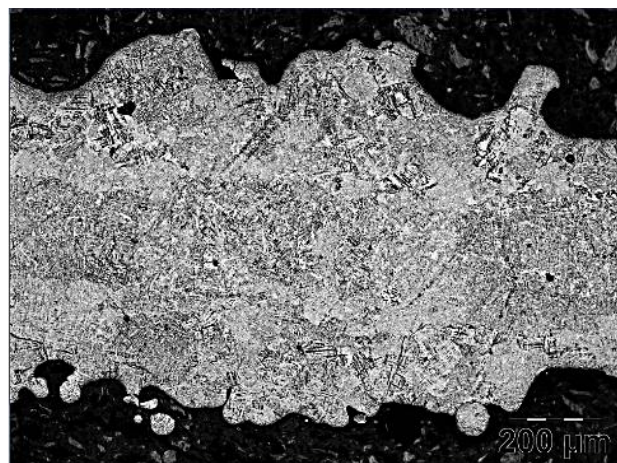


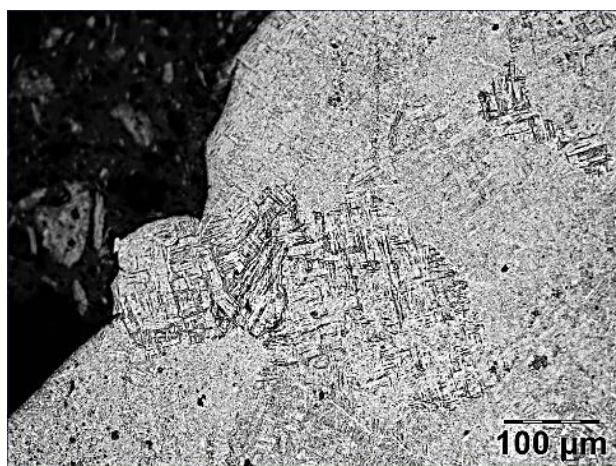
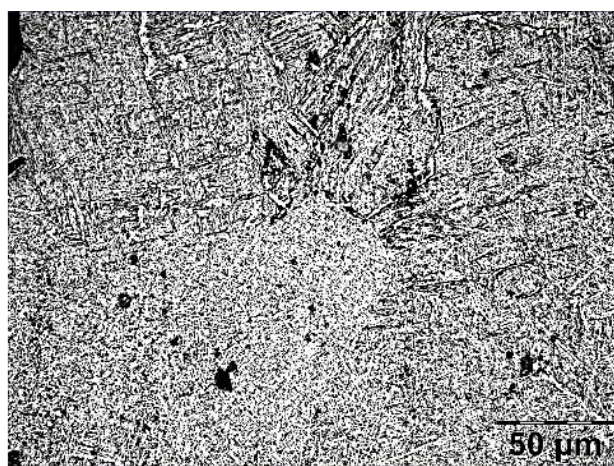
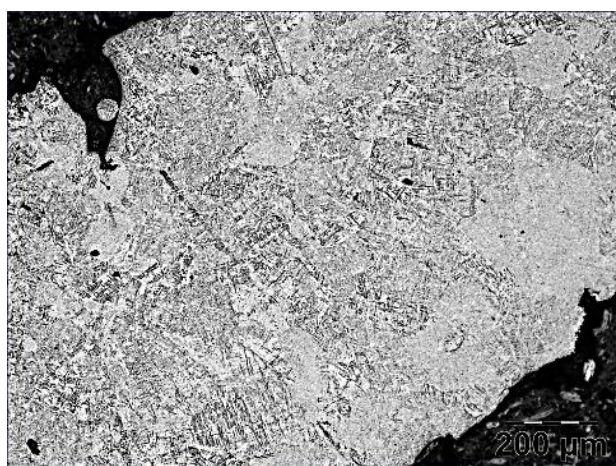
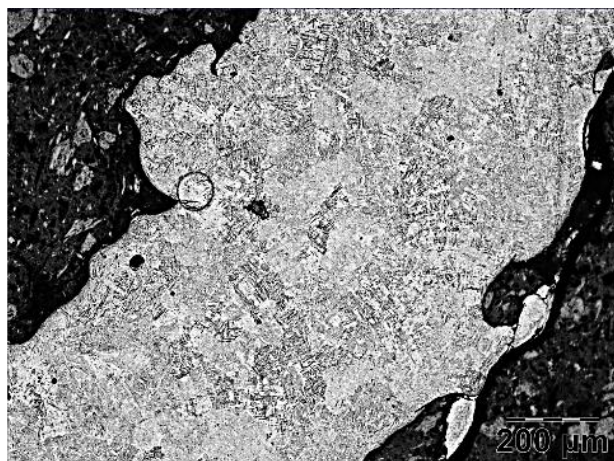
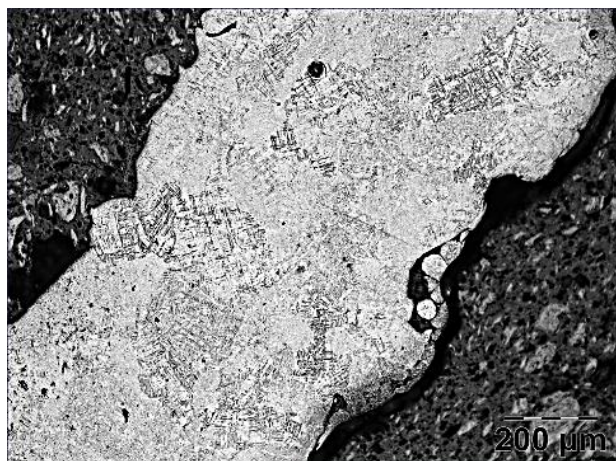


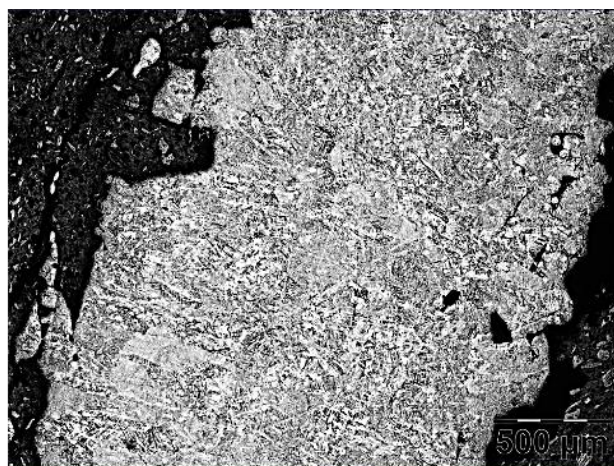
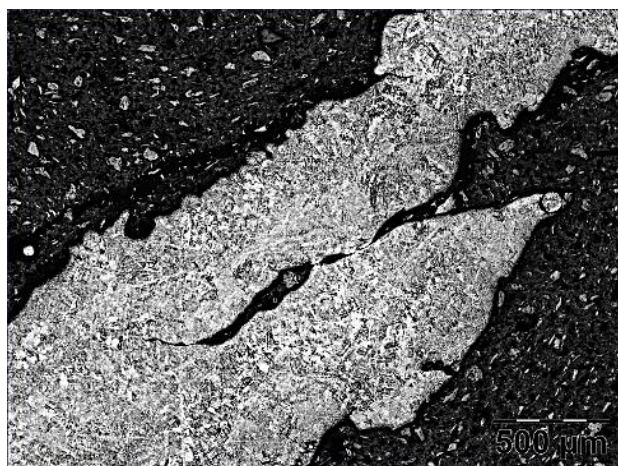
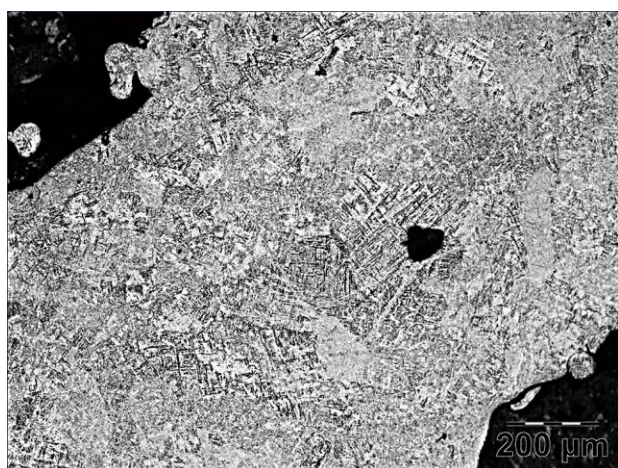
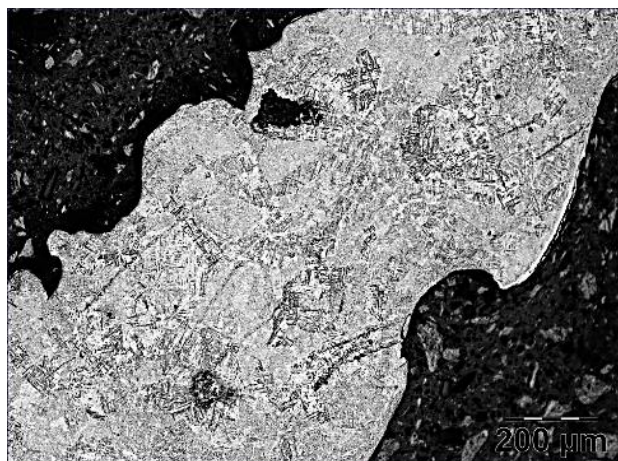


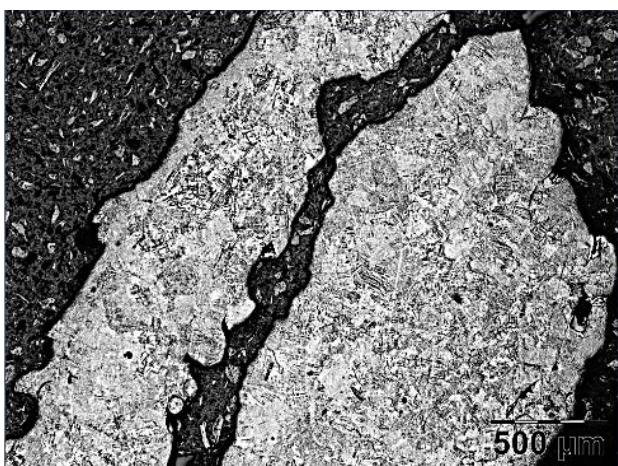
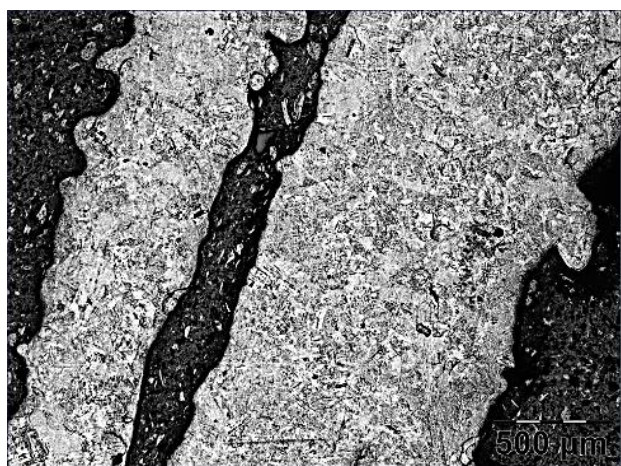
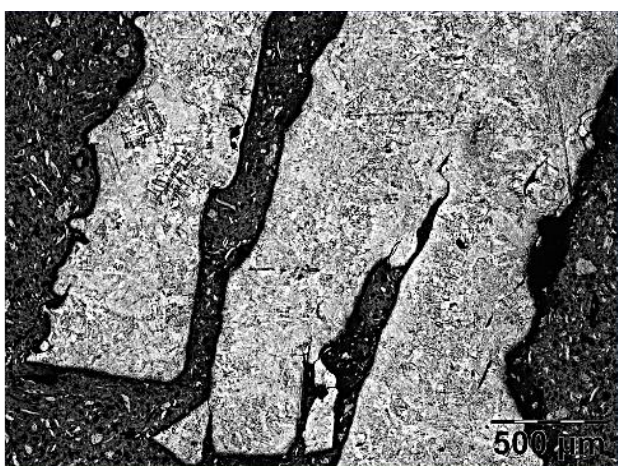
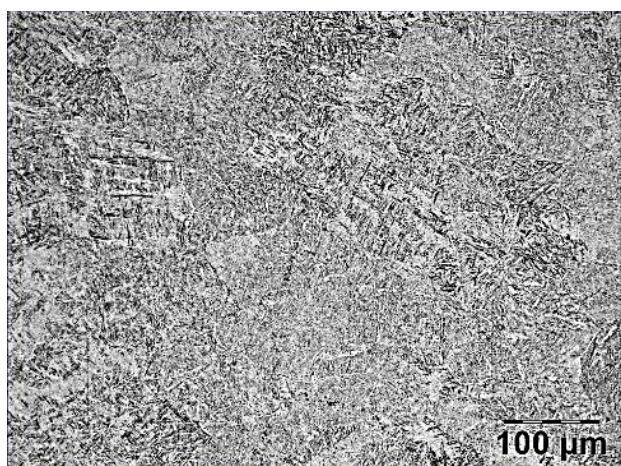
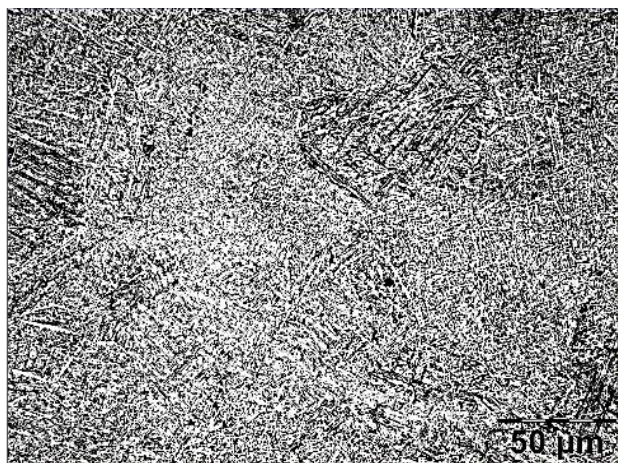
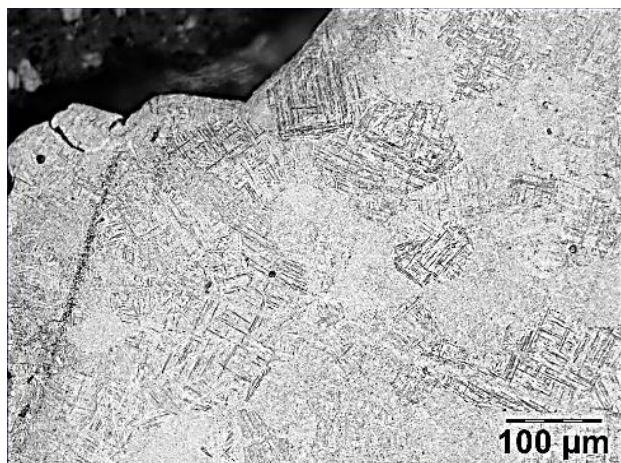
Specimen SV_D (Stress graded Diamond from Build Vertical) Build Direction in/out













VITA

Eric Bol has always been passionate for all things aviation and space related. From a young age he was a volunteer member of the Civil Air Patrol who rose to the rank of Cadet Major and served as a squadron commander at events. Originally, he targeted a military career in the US Army as a helicopter pilot with an ROTC scholarship; however, due to medical reasons he was disqualified from service by the Department of Defense. From there he pivoted to target an engineering career in aerospace, and in 2006, Eric received his Bachelor of Science in Aeronautical and Astronautical engineering from the University of Washington College of Engineering.

Following graduation Eric began his career at Boeing as a Flight Test Engineer. For five years he performed duties as a weights engineer and cabin supervisor on numerous test flights across all models in production (737, 747, 767, 777, 787). After first delivery of the 787-8 he transitioned to a design engineer within Propulsion Product Development. In that role his primary responsibility involved studying advanced nacelle architectures for inlets, fan cowls, thrust reversers, core cowls, plugs, pylons, aft fairings, and engine build up (EBU) systems. This is where Eric first learned about additive manufacturing and began investigating potential implementation of the technology in commercial aircraft propulsion products.

Eric Bol is currently an Associate Technical Fellow at Boeing with nearly 20 years of experience in the aerospace industry. As a propulsion system design and nacelle architect, he is working on developing new material systems and fabrication processes for future aircraft. While working full time, he attained his Master of Science degree in Mechanical Engineering in 2020 while pursuing his PhD through the Mechanical Engineering Department at the University of Washington. He has published multiple papers in highly regarded scientific journals, given many technical talks at industry conferences, and holds several patents. Through his research into metal additive manufacturing processes, he intends to lead technology development projects that will bring about the next step change in aircraft or spacecraft capability.