

**Relationship between Process, Microstructure, Design, and Properties
of Cast Aluminum Foams and Lattice Structures**

Von der Fakultät Georessourcen und Materialtechnik der
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Aachen, Jan. 2026

Mahan Firoozbakht

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0. Summary

0.1. Introduction

The exceptional properties of cellular materials, such as their lightweight nature, high energy absorption, and enhanced mechanical efficiency, have driven significant research interest in the development of porous structures across various material classes. Inspired by natural cellular architectures, researchers have aimed to fabricate foams with tailored properties for applications ranging from impact mitigation to thermal management. Among different cellular materials, polymeric foams have found widespread use in cushioning and packaging applications. However, the demand for stronger foams led to the exploration of metallic foams, which offer superior mechanical properties and durability compared to their polymeric counterparts.

Aluminum-based foams, in particular, have attracted considerable attention due to the excellent combination of low density and high specific strength offered by aluminum alloys. These materials exhibit significant potential for structural and functional applications in industries such as aerospace, automotive, and biomedical engineering. Various manufacturing techniques have been developed to produce aluminum foams, including powder metallurgy, gas injection, and space-holder methods. However, many of them present limitations in terms of pore structure control, scalability, and mechanical performance.

Investment casting has emerged as a promising alternative for the fabrication of open-pore aluminum foams and lattice structures, offering precise control over pore morphology and strut architecture through the use of lost patterns. This technique allows for the replication of complex porous templates, leading to highly customizable foam geometries with uniform pore distribution. Furthermore, investment casting enables the use of different aluminum alloy compositions, allowing for enhanced microstructural tuning and mechanical behavior optimization through post-processing treatments such as heat treatment and alloy modification.

Despite these advantages, challenges remain in optimizing the investment casting process, particularly with respect to producing defect-free struts, achieving homogeneous and refined microstructures, and tailoring mechanical properties at both the strut and foam levels. In this study, alterations were explored across different hierarchical levels. At the microstructural scale, coarse and fine features in stochastic aluminum foams were modified through alloying and casting parameters. At the structural scale, the alloying concepts developed for foams were transferred to lattice structures, where strut geometry (e.g., slenderness), additional load-

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bearing elements, and unit cell design were further varied. Figure 0-1 illustrates the hierarchical levels that influence the properties of open-cell foams. By combining alloy concepts with tailored lattice geometries, this work provides a comprehensive perspective on the interplay between processing, microstructure, design, and mechanical behavior as a final goal of the study.

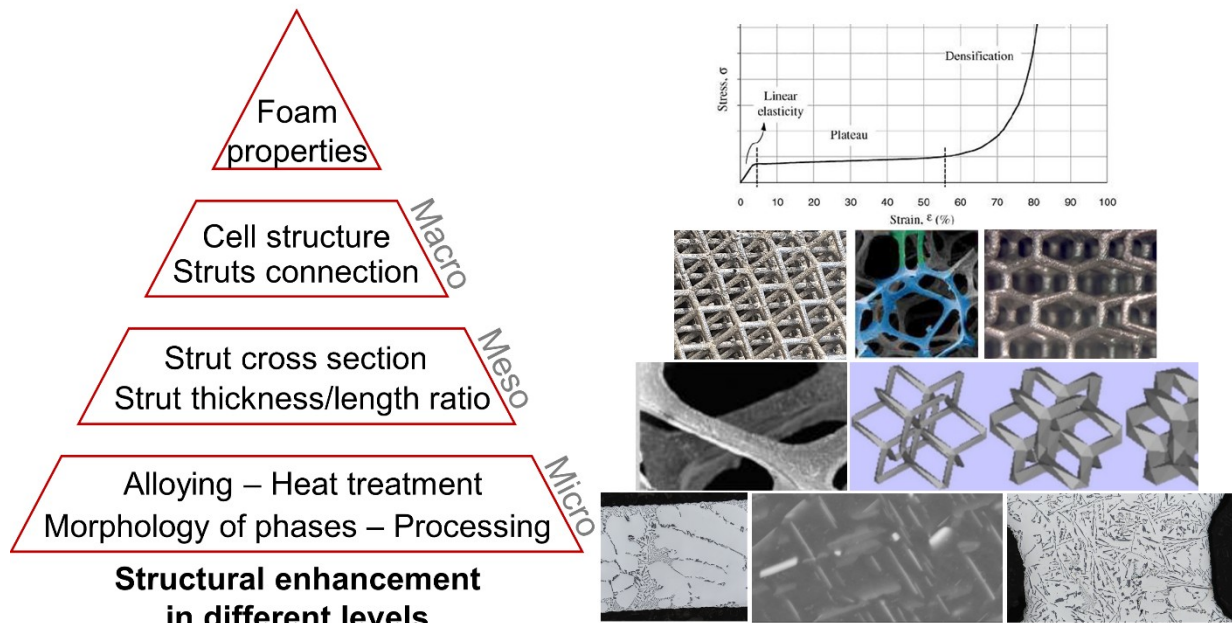


Figure 0-1- Different structural levels that influence the properties of open cell foams together with some practical examples.

0.2. Experimental work

Block mold casting, as a modified investment casting process, was used in this research to produce open-cell aluminum foams and lattice structures. This method allows for free design of the 3D geometry of the final structure, as the initial pattern dictates the foam's geometry and porosity profile. Two types of lost patterns were used: stochastic polyurethane (PUR) foams and 3D-printed resin preforms. These patterns were embedded in a ceramic slurry made from a gypsum-bonded investment material, which formed the mold for casting. A high mold temperature was maintained during casting to ensure proper mold filling, thereby reducing the risk of premature solidification and allowing for uniform melt distribution. Various aluminum alloys were cast, and the final cellular structures were evaluated for mechanical performance.

To characterize the mechanical properties of the cast structures, compression testing was conducted as a primary evaluation method. This provided insights into the deformation behavior, compressive strength, energy absorption efficiency, and load-bearing capacity of both stochastic foams and lattice structures. Additionally, individual struts were extracted from stochastic foams, and micro-tensile tests were

performed to assess their tensile properties in relation to the microstructural variations. This combination of mechanical testing approaches facilitated a comprehensive understanding of how alloy composition and casting conditions influence both bulk and localized mechanical responses in the produced cellular materials.

0.3. Result and discussion

Evaluation of various casting conditions, particularly the effects of mold and melt temperatures, has identified a specific process window for producing stochastic metal foams with complete mold filling. Achieving complete mold filling is essential for maximizing the relative density of the foam, which in turn improves the mechanical properties. A mold temperature higher than the liquidus temperature of the melt ensured full mold filling and a consistent solidification process. Under these conditions, nucleation begins within the struts as endogenous solidification, progressing towards the mold surface. As solidification continues, a single aluminum grain forms in a dendritic structure in each strut, while most eutectic silicon particles accumulate on the strut surface.

Decreasing the mold temperature to the range between the liquidus and solidus temperatures of the melt results in a transition to a more favorable microstructure, characterized by globular aluminum grains and a more homogeneous distribution of the silicon phase. While this adjustment is expected to enhance the overall quality of the foam, the reduction in mold filling and the production of thinner struts presented debatable considerations.

Observation of foam deformation and strut tensile performance revealed a more brittle behavior in foams with a dendritic aluminum structure compared to those with a multi-grain globular microstructure. In the dendritic structure, the accumulation of silicon particles near the strut surface, along with sharp strut edges, led to brittle strut fractures with clear indications during the foam compression test as sudden stress drops after peak stress and fluctuations in the plateau region. By contrast, the globular microstructure, characterized by a more uniform silicon distribution, showed ductile deformation under compression, with smoother strain progression and reduced stress fluctuations. Despite this improved ductility, the globular microstructure exhibited a lower overall stress tolerance than the dendritic counterpart. This was attributed to thinner struts, resulting from reduced mold filling at the lower mold temperatures required to achieve the globular grain structure. While the microstructure was more favorable, the decrease in mechanical performance emphasizes the complex interplay between processing conditions, microstructure, and strut geometry in determining foam performance. Average

0. Summary

compression stress–strain curves of foam samples for two casting conditions that resulted in dendritic grain structures and one condition that resulted in a globular grain structure are shown in Figure 0-2.

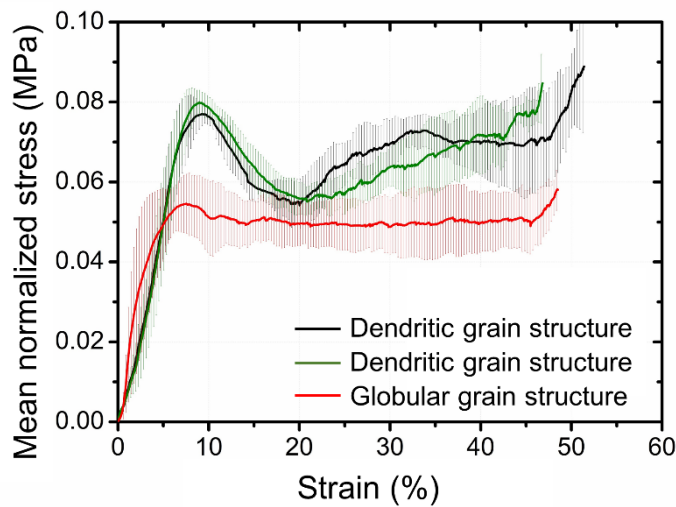


Figure 0-2- Mean normalized stress curves for two casting conditions resulted in dendritic grain structures and one casting condition resulted in a globular grain structure.

While solidification predominantly governs the coarse features of the microstructure, such as grain morphology and eutectic phase distribution, non-uniform cooling during block mold casting plays a key role in shaping finer microstructural details. Temperature measurements during casting revealed different cooling rates in different sections of the foam samples: areas near the pouring basin or surrounded by more molding compound exhibited slower cooling. Although the variations in cooling rate were moderate compared to other casting methods, they had a pronounced impact on finer microstructural features, such as the transformation of the β -phase into globular intermetallics in faster-cooled regions. These findings highlight the high sensitivity of foam castings to local cooling conditions, especially when casting geometry is restricted and solidification dynamics vary locally.

An important aspect of fine-tuning the microstructure of Al-Si alloys is the modification of silicon particles, which is particularly crucial in aluminum foams with slow cooling rates during solidification, resulting in coarse silicon particles. The effectiveness of chemical modifiers such as strontium under these conditions must therefore be carefully evaluated. To study this, radial foam samples were positioned along the radial direction of the flasks, generating a cooling gradient ranging from approximately 0.5 °C/s near the flask wall to 0.1 °C/s at the flask center. Sr addition proved effective in modifying the silicon phase in regions with higher cooling rates near the wall as shown in Figure 0-3, while its influence diminished toward the center where cooling was slower. The influence of strontium on the mechanical properties of Al-Si foams was evident at both the foam and strut levels. At the foam scale, Sr addition improved the uniformity of stress–strain

responses, reduced fluctuations during deformation, and promoted a more ductile behavior with enhanced energy absorption efficiency. At the strut level, the addition of 200–400 ppm Sr increased both tensile strength and elongation at break, indicating improved material ductility and load-bearing capacity. However, when the Sr content was increased to 800 ppm, a decline in mechanical performance was observed, likely due to the over-modification effect and the formation of deleterious intermetallic phases.

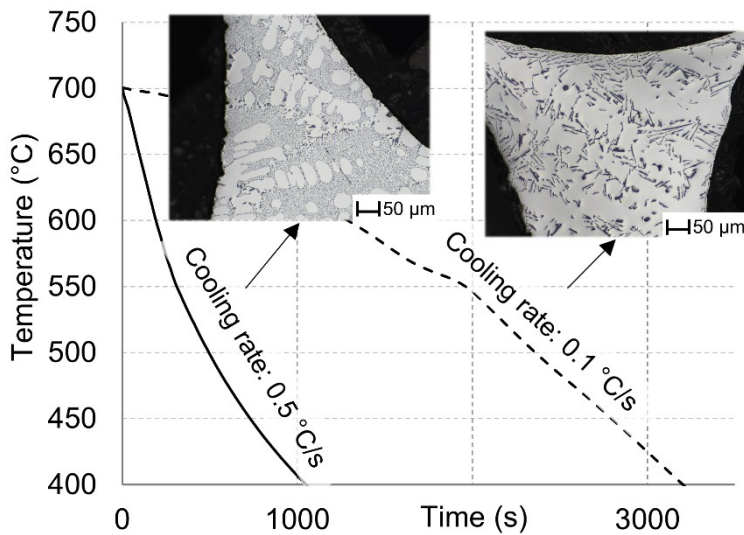


Figure 0-3- Variation of cooling rate in the horizontal sample, decreasing from 0.5 °C/s near the flask wall to 0.1 °C/s at the center, and the corresponding microstructural changes in Al10SiMg alloy with the addition of 400 ppm Sr.

In contrast to Al-Si alloys, where microstructural refinement is largely dependent on eutectic modification, Al-Cu alloys offer an alternative way of customizing properties through heat treatment. Precipitation of copper-rich particles in the Al-Cu alloy system through heat treatment offers a practical means to create size-adjusted reinforcing phases. To fully benefit from this mechanism, it is essential to perform an effective solution annealing treatment. This step helps to homogenize the microstructure by redistributing segregated elements and dissolving coarse non-equilibrium phases that form during solidification. When followed by controlled aging, this process enables the formation of fine, well-dispersed precipitates, which play a key role in strengthening the struts of open-cell foams. Aging treatments at 160 °C and 200 °C resulted in the formation of fine precipitates, likely GP zones or θ' phase, that enhanced strut hardness and foam strength, while aging at 300 °C led to coarse θ precipitates and performance similar to the as-cast condition. In the absence of brittle phases, deformation proceeded in a stable and ductile manner, as evidenced by smooth stress–strain curves. The homogeneous microstructure of Al-Cu foams enabled a direct correlation between intrinsic material properties, such as hardness, and the macroscopic mechanical response.

To move beyond the general design of the stochastic open-cell foams and systematically investigate the influence of structural design on mechanical

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performance, a process chain integrating additively manufactured sacrificial patterns into the block mold casting was established. This hybrid approach enabled simultaneous exploration of microstructural and geometrical (macrostructural) optimization.

Lattice structures with BCC, Diamond, and Kelvin topologies were fabricated via VAT photopolymerization and subsequently cast using the AlSi10MgMn alloy. Despite maintaining identical relative densities across the designs, each lattice type exhibited distinct deformation behavior as in Figure 0-4. The BCC structure showed localized shear band formation and brittle failure tendencies, Diamond offered superior load-bearing capacity due to its short struts and internal complexity, and Kelvin demonstrated the most ductile response, resulting in high energy absorption. Although the designed lattices were expected to deform in a bending-dominated mode, thick struts and casting-related issues led to abrupt failures and fluctuations in the stress–strain response.

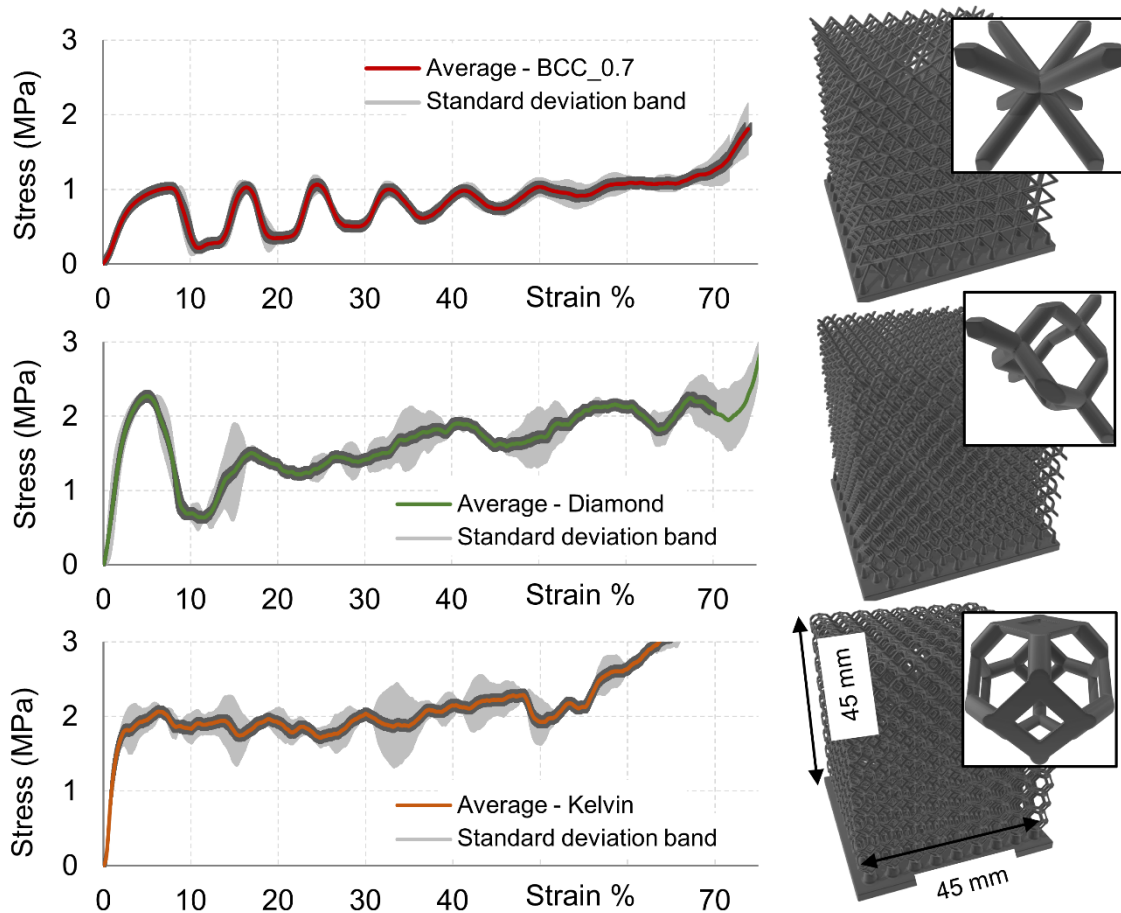


Figure 0-4- Average stress strain curves of lattice structures with BCC, Diamond, and Kelvin unit cell designs at the same relative density.

In comparison, BCC lattices produced by LPBF under optimized parameters exhibited fully ductile deformation, attributed to the fine microstructure and the

beneficial effect of post-processing treatments like stress-relief annealing. However, achieving optimal properties via LPBF requires careful process control and material availability, which can be time-consuming and costly. Thus, the hybrid investment casting approach provides a practical and efficient platform for initial design and material evaluation, offering valuable insights into the deformation behavior of complex lattice structures before transitioning to direct additive manufacturing for final part production.

Further optimization efforts focused on the BCC lattice structure, which exhibited a higher susceptibility to brittle deformation and therefore provided a more suitable case for studying the influence of material and structure modifications. To investigate the role of ductility, the silicon content in the Al-Si alloy system was varied (3, 7, and 10 wt%) in BCC structures. As expected, shifting the alloy toward a more brittle behavior resulted in higher stiffness and peak stress, but also led to more frequent occurrences of sudden strut fractures and pronounced serrations in the stress-strain curves. Additionally, although brittle fractures occurred at earlier strain levels, the overall densification of the structure was delayed, indicating that stronger and stiffer struts were still able to resist collapse over a longer deformation path, despite the presence of local failure events.

To further investigate the role of microstructural refinement in lattice structures, the effects of Sr addition and the creation of a globular aluminum grain structure, achieved by casting in molds with lower temperatures, were examined. While these modifications were promising for stochastic foams with thin struts that can bend easily, they did not significantly affect the mechanical properties or deformation behavior of BCC lattice structures with thicker struts. The results indicate that microstructural refinement alone is insufficient to improve ductility in cast lattice structures, particularly when deformation is governed by localized fracture mechanisms instead of bending-driven failure modes. The dominant factors influencing failure in these thicker-strut lattices are more related to strut geometry, stress distribution, and the presence of defects, rather than intrinsic ductility.

To investigate the influence of structural changes, variations in strut thickness and the addition of extra struts to create stretch-dominated structures were investigated. Increasing the slenderness ratio allowed for greater bending deformation, which in turn favored ductile behavior. This observation supports the idea that microstructural improvements can be used more effectively when the struts are capable of bending, so that internal structural refinements can contribute to a more favorable deformation mechanism. When additional struts were introduced into the BCC structure, their orientation played a critical role. Struts aligned with the loading direction significantly enhanced mechanical performance,

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as they directly contributed to load-bearing. In contrast, struts oriented perpendicular to the loading axis provided little to no improvement. In these cases, high node spacing and the inherent limitations of the casting microstructure prevented the added struts from effectively bearing tensile loads. This highlights a key constraint in hybrid investment-cast structures: while compressive loads can be absorbed efficiently by structural elements, tensile load-bearing elements remain a challenge.

Finally, combining alloy modification with an improved structural design that incorporated additional vertically aligned struts demonstrated that the most ductile alloy (AlSi3) not only provided greater stability, with reduced stress fluctuations, but also achieved higher post-yield strength (Figure 0-5). In this configuration, the added struts deformed plastically, thereby contributing to the mechanical performance throughout the entire compression process.

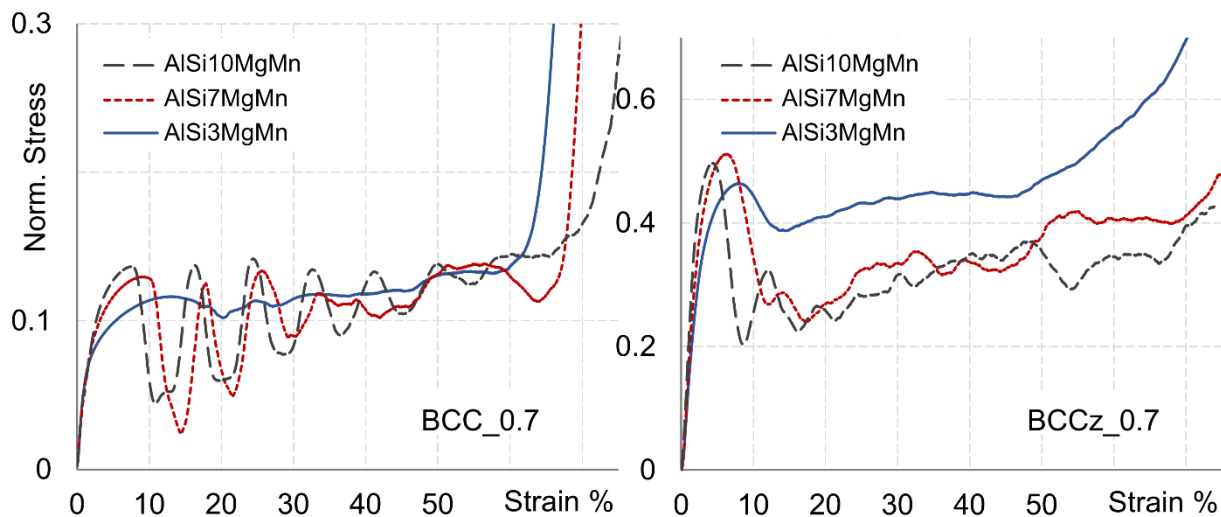


Figure 0-5- Normalized stress-strain curves for BCC structures (strut diameter of 0.73 mm) without (BCC_0.7) and with extra struts in the vertical direction (BCCz_0.7).

The investigations conducted in this work provide a comprehensive analysis of the interplay between microstructure, geometry, and mechanical behavior in open-cell aluminum cellular structures produced by investment casting. It was shown that confined solidification spaces, such as the micrometer-scale struts in foams, exhibit higher sensitivity to local cooling rates, enabling microstructural changes even under overall slow cooling conditions. While such microstructural improvements effectively enhanced the mechanical properties of foams, they were insufficient to prevent brittle failure in lattice structures. In these cases, base alloy selection and geometrical design modifications had a more pronounced impact.

1. Introduction and Motivation

Metallic lattice structures with complex cell geometries and hollow inner architectures have been the focus of research activities for several years. They represent exceptional materials for lightweight construction concepts, capable of withstanding high loads and absorbing energy in the event of an impact. These porous metal structures are considered the next generation of metallic foams, distinguished by entirely different geometrical arrangements [1–4]. While foams exhibit stochastic arrangements, lattice structures are ordered and formed through the repetition of a primary cell geometry. The ultimate mechanical properties of metal foams and lattice structures can be influenced at various levels, namely by the microstructure, the cell design, and the interconnection of the cells [5–23].

Due to the complexity of the geometry, sophisticated processing methods such as additive manufacturing are ideal for the production of lattice structures [24–33]. In the case of direct additive manufacturing, the finished metal part is produced by metal 3D printing, primarily through layer-by-layer sintering of metal powder. Indirect additive manufacturing, on the other hand, is a multi-step production process in which sacrificial preforms, replicating the final geometries, are produced by additive manufacturing in the first step [34–43]. These preforms are then used as lost patterns or lost molds in the final casting process. Owing to differences in cooling behavior and the fundamental principles of these two process routes (direct and indirect additive manufacturing), the resulting microstructure and, consequently, the mechanical properties of the structures produced by each method are markedly different. In direct additive manufacturing, the melt pool is small and solidifies rapidly due to high heat extraction, leading to a very fine microstructure. In contrast, the cast microstructure is dendritic and coarse, and does not even resemble that of conventional casting processes, as the solidification space is constrained by the geometry of the strut [44–47]. Although many studies have focused on the effects of the production process on the properties of printed lattice structures, comparatively less attention has been paid to the cast structures. In casting, variations in the alloy and some process parameters can be achieved more easily, and the extensive historical knowledge of casting can be employed to tailor the microstructure and mechanical properties of lattice structures.

In this regard, aluminum foams and lattice structures are particularly significant among metallic cellular materials due to their combination of material properties, processing versatility, and broad application potential [11,48–50]. Aluminum is inherently lightweight, and incorporating it into a cellular structure further enhances its strength-to-weight ratio, making it ideal for applications where weight

reduction is critical without compromising strength. This property is essential in the aerospace, automotive, and transportation industries, where minimizing weight directly contributes to fuel savings and improved efficiency. The diverse structural and functional applications of cellular aluminum structures have attracted significant research interest and opened up substantial opportunities for innovation and advancement across various fields.

Based on this foundation, the current dissertation aims to deepen the understanding of aluminum cellular structures by examining how the investment casting process and alloy composition influence deformation behavior and critical mechanical properties. Furthermore, this study advances fundamental knowledge of cellular material design and provides a pathway for optimizing manufacturing processes and alloy properties to meet the evolving demands for these lightweight yet durable materials. After presenting the current state of the art, the most significant influencing parameters of block mold casting and their effects on the solidification and microstructure of cast aluminum foams are discussed. By varying the main alloying elements, the general mechanical properties are altered to observe their impact on deformation behavior. Further microstructural modifications are also targeted and analyzed to investigate the impact of finer changes. For these investigations, stochastic foam samples are initially produced, and their microstructure and mechanical properties are evaluated. Subsequently, different cell designs are employed to explore the relationship between the alloy concept, the resulting microstructure, and the deformation response of different lattice structures. A comparison of different production routes of direct and indirect additive manufacturing is also conducted for one lattice structure design.

By analyzing how these factors affect the microstructure and overall performance of aluminum foams and lattice structures, this research seeks to provide insights into optimizing their properties through different structural levels. The findings are anticipated to contribute to more efficient design and production of aluminum cellular materials, enhancing their versatility and functionality across a range of applications.

2. State of the Art

2.1. Cellular solids

Long-term studies of natural structural materials have demonstrated that many of them are based on cellular structures rather than dense solids [5,51]. Nature often relies on sophisticated cellular architectures to achieve strength and durability while minimizing weight. For example, materials such as wood, bone, and coral utilize a network of cells to provide structural integrity. These natural cellular structures are highly efficient, requiring only minimal material to withstand external forces, making them effective for various biological functions as well. This natural blueprint has inspired scientists and engineers to replicate similar structures in synthetic materials, leading to the development of engineered cellular solids [50,52–54].

Cellular solids are characterized by a porous structure, in which a solid framework of interconnected struts or plates surrounds air-filled cavities [51]. This configuration imparts unique properties such as low density, high strength-to-weight ratio, and excellent energy absorption capabilities. The geometry and interconnection of the cells play a crucial role in determining the mechanical, thermal, and functional properties of these materials [5,9]. As a result, cellular solids can be tailored for a wide range of applications, offering levels of customization that dense materials cannot easily achieve.

Due to their versatile properties, cellular solids have found applications across multiple industries, ranging from aerospace and automotive to biomedical and construction sectors [1,55]. Among the most promising engineered cellular materials are aluminum foams and aluminum lattice structures. They combine the lightweight and energy-absorbing benefits of cellular structures with the strength and thermal conductivity of aluminum, making them ideal for applications such as impact protection and heat management solutions [1,3]. On the other hand, aluminum lattice structures, with their ordered networks of interconnected beams or nodes, offer precise control over mechanical properties with high design flexibility. These cellular structures can be engineered to optimize load-bearing capacity, stiffness, and energy absorption, providing tailored solutions for advanced structural applications. The following sections explore the applications, structure, properties, and production of these innovative materials.

The diversity of metal cellular materials can be primarily categorized into three types: open-cell foams, closed-cell foams, and lattice structures, as shown in Figure 2-1. Each type exhibits a distinct internal architecture that significantly influences

its mechanical behavior and suitability for specific applications. Understanding the differences between these types of cellular solids is essential for selecting the appropriate material for a given engineering task.

Open-cell metal foams consist of a porous network of interconnected voids, resembling a three-dimensional web of thin struts. This structure allows fluids, such as air or liquids, to pass through the foam, making open-cell foams ideal for applications like filtration and heat exchangers [56]. However, their open structure can limit their structural strength compared to closed-cell foams, making them more suitable for applications where fluid flow or high energy absorption efficiency are essential [1].

Closed-cell metal foams, on the other hand, consist of isolated, gas-filled cavities surrounded by solid walls. This configuration provides higher compressive strength and rigidity compared to open-cell foams, as the closed-cell structure resists deformation more effectively under load [57]. Closed-cell foams are well-suited for lightweight structural components, insulation, and impact protection, where both low density and mechanical integrity are required [20,58]. Additionally, the closed nature of the cells enables these foams to float, making them useful in applications such as marine structures and buoyancy aids [51].

Lattice structures represent a more ordered and precise approach to cellular architecture [59]. Unlike traditional foams, which feature a more random cell distribution, lattice structures consist of carefully designed networks of cells. These structures can be produced using advanced manufacturing techniques like additive manufacturing, which allows for precise control over cell size, geometry, and density. This capability enables engineers to tailor lattice structures to achieve specific mechanical properties, such as optimized strength-to-weight ratios and customized deformation behavior [60]. As a result, lattice structures are increasingly used in high-performance applications.

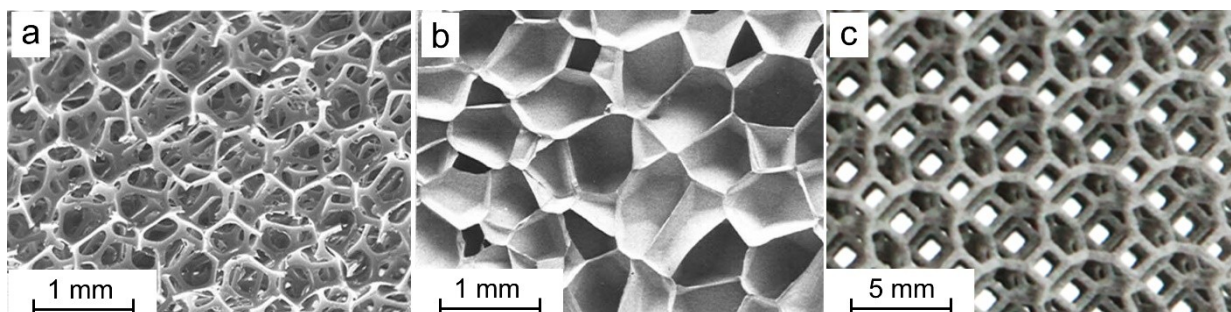


Figure 2-1- Classification of cellular solids in a) open-cell, b) closed-cell, and c) lattice structure [5].

2.2. Applications

2.2.1. Mechanical and structural applications

Cellular solids are ideal for applications where mechanical strength, energy absorption, and structural integrity are required [61]. Their ability to withstand loads while remaining lightweight makes them particularly suitable for industries such as automotive, aerospace, and construction. One of the most prominent mechanical applications of cellular solids is in energy absorption systems [58,62]. Open-cell and closed-cell foams are commonly used in automotive crash pads, bumpers, and side-impact protection components [19,58]. These materials can absorb significant amounts of energy during collisions, helping to minimize the impact forces transmitted to passengers and improving safety. The porous structure allows the foam to collapse gradually, dissipating energy and preventing sudden deceleration.

In addition to energy absorption, cellular solids are widely used in load-bearing and structural applications where their strength-to-weight ratio provides a significant advantage [1]. In aerospace, lightweight lattice structures and aluminum foams are employed in structural panels, fuselage components, and other load-bearing parts. These materials contribute to reducing the overall weight of aircraft, thereby improving fuel efficiency without compromising the structural integrity needed to withstand the stresses of flight. Similarly, in the construction industry, cellular solids are utilized in structural panels, beams, and floor systems, where both low weight and high compressive strength are critical.

2.2.2. Functional applications

Beyond their structural applications, cellular solids are also valued for their functional properties, including permeability, thermal conductivity, and acoustic performance [2,61,63]. These characteristics make them well-suited for applications in thermal management, filtration, and sound absorption.

Open-cell foams are particularly effective as filters and heat exchangers due to their interconnected pore networks [56,64]. The porous structure of these materials allows fluids, such as air or liquids, to flow through easily, making them ideal for filtration systems that require a high surface area and minimal pressure drop. In thermal management applications, open-cell foams function as efficient heat exchangers or cooling elements [65]. Their high surface area enables rapid heat dissipation, making them suitable for cooling systems in electronic devices, engines, and other equipment where thermal regulation is critical. The controlled porosity and permeability of these materials enhance heat transfer between fluids, improving overall cooling performance. Additionally, closed-cell foams are

commonly used as thermal insulation materials in building construction, where their resistance to heat transfer contributes to energy efficiency.

The biomedical field uses the porous nature of cellular solids for applications such as bone implants and scaffolds for tissue engineering [66,67]. Open-cell metal foams, particularly those made from biocompatible materials like titanium or aluminum, are increasingly used as bone grafts due to their ability to facilitate bone in-growth. Their interconnected pore structure promotes better integration with natural bone tissue, aiding in healing and reducing the risk of implant rejection. Furthermore, lattice structures, designed with precise geometries, can be used to tailor the mechanical properties of implants, ensuring the right balance between strength and flexibility to match the characteristics of natural bone.

2.3. Effective structural parameters

The performance and mechanical behavior of cellular solids are strongly influenced by their structural parameters. Critical factors, such as pore size, strut or wall thickness, and relative density, directly affect key properties [3,5]. A thorough understanding of these parameters allows engineers to customize cellular materials for specific applications, ensuring an optimal balance between weight and performance.

2.3.1. Pore size and shape

The size and shape of pores within cellular solids are crucial factors in determining their mechanical and functional properties [13,52]. Pore characteristics significantly influence properties such as compressive strength, energy absorption, thermal conductivity, and fluid permeability. These parameters can be precisely tailored during the manufacturing process to suit the specific needs of various applications. The effects of pore size and shape differ depending on whether the foam is of the open-cell or closed-cell type, as each structure interacts with external loads and environmental factors in different ways.

In open-cell foams, the network of interconnected voids creates a permeable structure, and the pore size significantly impacts several key properties. Larger pore sizes enhance permeability, making these foams especially useful in applications such as filters and heat exchangers, where fluid or gas flow through the material is essential [3,64]. In these cases, larger pores facilitate lower pressure drops and improved flow rates. However, larger pores may also reduce the overall compressive strength and stiffness of the foam, as the solid framework is spread over a larger void space, making the material more susceptible to deformation. In contrast, smaller pores result in a denser structure with increased surface area, which can improve energy absorption and provide better resistance to compression.

[11]. This makes open-cell foams with smaller pores well-suited for cushioning, impact absorption, and damping applications, where a balance between energy absorption and structural integrity is critical.

In closed-cell foams, the pores are encapsulated by solid walls, creating a structure where each void is isolated from its neighbors. The pore size in closed-cell foams has a direct impact on properties such as buoyancy, insulation, and compressive strength [9]. Larger pores generally reduce the foam's density, making the material lighter and more buoyant, which is beneficial for applications like flotation devices and lightweight fillers. However, this also means that the walls surrounding each cell are thinner, potentially reducing the foam's strength and making it more prone to localized buckling or collapse under compressive loads.

Conversely, smaller pores create a more compact arrangement of gas-filled cells, resulting in a foam that is denser and stiffer. The thicker cell walls provide better support under compression, enhancing the foam's structural performance and making it suitable for load-bearing applications, such as lightweight structural panels [68]. Smaller pores also improve thermal insulation by suppressing gas convection within individual cells and limiting radiative transfer. These factors make closed-cell foams with smaller pores particularly effective for applications where thermal management or insulation is a priority.

In both open- and closed-cell foams, the shape of the pores is another important factor [21]. Pores can be spherical, polyhedral, or irregularly shaped, and this geometry affects how the foam deforms under external forces. Spherical or near-spherical pores tend to distribute stress more evenly throughout the foam, resulting in more isotropic behavior and higher compressive strength [51,69]. In contrast, irregular or polyhedral pores can introduce anisotropic properties, meaning that the material's strength and stiffness vary depending on the direction of the applied load [6]. While this can be a limitation in some cases, it can also be beneficial when specific directional properties are required [12]. For instance, the Young's modulus, as well as the elastic and plastic collapse stresses, are greater in the rise direction than in the normal direction of open-cell foams [6,9].

2.3.2. Strut/wall thickness

The thickness of the struts or walls that form the cellular framework of metal foams is another significant influencing factor. In both open-cell and closed-cell foams, thicker elements provide greater structural integrity and resistance to deformation under compressive loads [51]. However, while increasing strut/wall thickness can enhance strength, it may also lead to an increase in weight. Therefore, a careful balance must be struck between the desired mechanical properties and the overall

weight of the material, particularly in industries where weight reduction is critical. In applications where lower density is desired, thinner struts may be used, but this often comes at the expense of strength and stability.

Strut thickness also affects the energy absorption characteristics of cellular solids. In open-cell foams, thinner struts can improve the foam's ability to compress and deform under impact, enabling more gradual energy dissipation during collisions or dynamic loading [70]. This behavior is particularly valuable in applications like crash pads or protective gear, where the primary goal is to absorb and dissipate energy while minimizing injury. However, the trade-off is that thinner struts may reduce the overall strength and stability of the foam, making it more susceptible to failure under high loads.

2.3.3. Relative density

Relative density is a critical structural parameter in the study of cellular solids, defined as the ratio of the density of the cellular material to the density of the solid material from which it is derived [5,9,51]. This parameter significantly influences the mechanical properties of cellular structures, including strength, stiffness, and energy absorption [71]. Generally, as relative density increases, the strength and stiffness of the material also improve. A higher relative density indicates a greater amount of solid material in a given volume, leading to enhanced load-bearing capacity and resistance to deformation. In contrast, lower relative density typically results in lighter materials with higher energy absorption capacity. Cellular solids with low relative density can deform more easily under load, facilitating greater energy dissipation during impacts.

Several empirical and semi-empirical models have been developed to relate the mechanical properties of cellular solids to their relative density. One widely used approach is the power law [51,70,72], which characterizes how certain mechanical properties scale with relative density and is further discussed in the next section.

2.4. Mechanical properties

2.4.1. Structural hierarchy in the cellular materials

Many natural and man-made materials have a structure on multiple length scales, with the structural elements themselves having their own internal hierarchies [7]. This structural hierarchy plays a significant role in determining the material properties. In cellular materials, the overall properties are influenced by a hierarchical structure that can be described across three primary levels: micro-level, meso-level, and macro-level [53]. Each of these levels uniquely contributes to the material's overall performance. For example, in open-cell metallic cellular structure, these levels and the factors influencing them are illustrated in Figure 2-2.

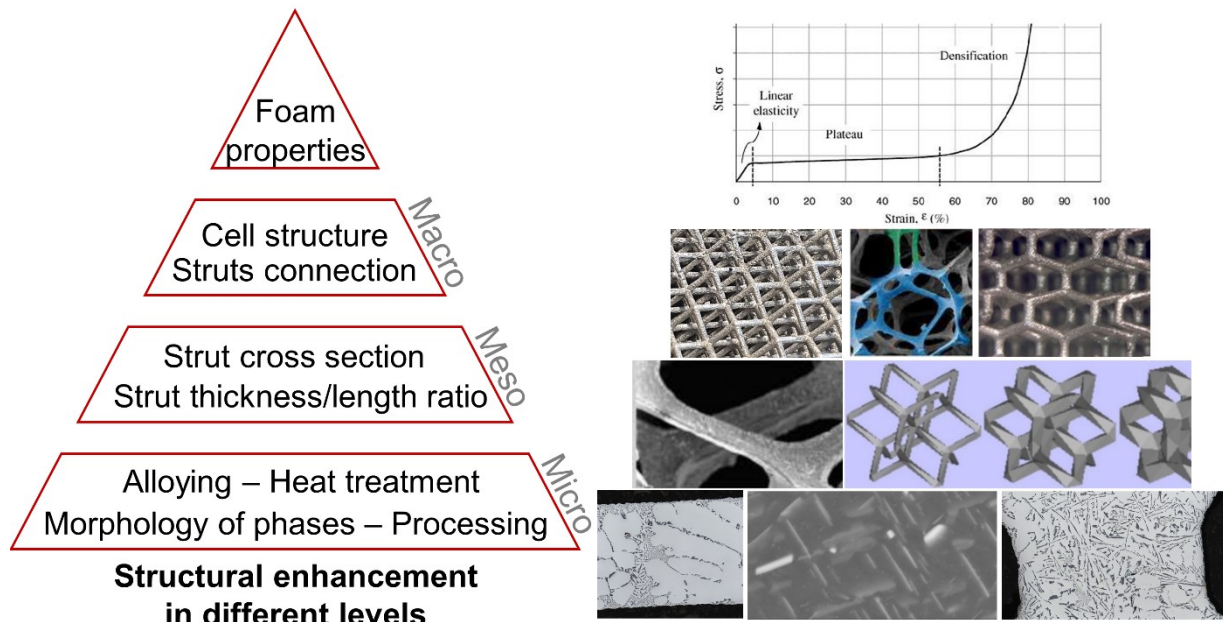


Figure 2-2- Effective structural hierarchy of metallic cellular materials, highlighting the Micro-, Meso-, and Macro-levels, along with examples of influencing parameters at each level for the open-cell foam.

Micro-level: At the micro-level, the focus is on the intrinsic material properties within the cell walls or struts [53]. This includes the grain structure, phase distribution, and other material characteristics within the solid matrix. Microstructural elements, such as grain size and phase distribution, play a critical role in determining key mechanical properties like yield strength, fracture behavior, and fracture toughness [12]. These intrinsic properties can be modified through alloying, processing, and heat treatment [71], and they significantly influence how the cellular material performs under stress and impact, forming the fundamental basis for the overall material performance.

Meso-level: The meso-level refers to the cellular architecture, focusing on the characteristics of individual cells and their components, such as struts or cell walls [53]. Key aspects at this level include pore size, cell shape, strut/wall geometry and thickness, as well as whether the structure is open- or closed-cell. These parameters control the material's density, porosity, and deformation behavior, which directly influence its bulk properties.

Macro-level: At the macro-level, the structure is considered as a whole, encompassing the overall shape, size, and orientation of the cellular material within its application context [53]. Macrostructural features include the arrangement of cells and their connections within a larger component, variations in cell density across regions, and the directional orientation of cells to match expected loading

conditions [9]. The macro-level design is crucial for ensuring the material performs optimally under application-specific stresses, such as bending, compression, or impact. For instance, cellular materials designed with gradient density at the macro-level can enhance impact resistance by concentrating higher density in critical areas, while the combination of failure modes provides benefits in energy absorption [62,73].

2.4.2. Mechanical response under compression

Based on the structural applications of cellular materials, they are primarily stressed under compression. However, during compression, various structural elements (such as struts or cell walls) in different directions may experience different stress states, including compression, tension, bending, or torsion. Schematic stress-strain curves for foams exhibiting different behaviors under uniaxial compression are shown in Figure 2-3. At low strains, the foam deforms in a quasi-elastic manner. This is followed by a plateau region of almost constant stress, and finally, a densification region. The extent and characteristics of each region depend on the interrelation of the structural levels. Elastic, plastic, and even brittle foams all exhibit three-part stress-strain curves, although the mechanisms driving each step differ.

Quasi-elastic region: Initially, when the load is applied, the cell struts/walls bend elastically, which is limited to small strains. Upon stress release, the deformation may not fully restore in a linear manner due to the buckling of the cells [5]. Therefore, this region is called quasi-elastic, as it resembles the elastic behavior of bulk materials under deformation, though it is not identical to the behavior observed in solid samples. The slope of this section of the stress-strain curve is referred to as the elastic gradient or, more similarly to bulk materials, the Young's modulus (E). This gradient characterizes the foam's stiffness and can be related to its relative density, as shown in Equation 1 [5,9,70].

$$\frac{E}{E_s} = C \cdot \rho_r^m \quad \text{Equation 1}$$

where C is a material constant, E_s is the elastic modulus of the base material, ρ_r is the relative density, and m typically ranges from 2 to 3 for cellular materials. As previously mentioned, this relationship signifies that stiffness increases with relative density.

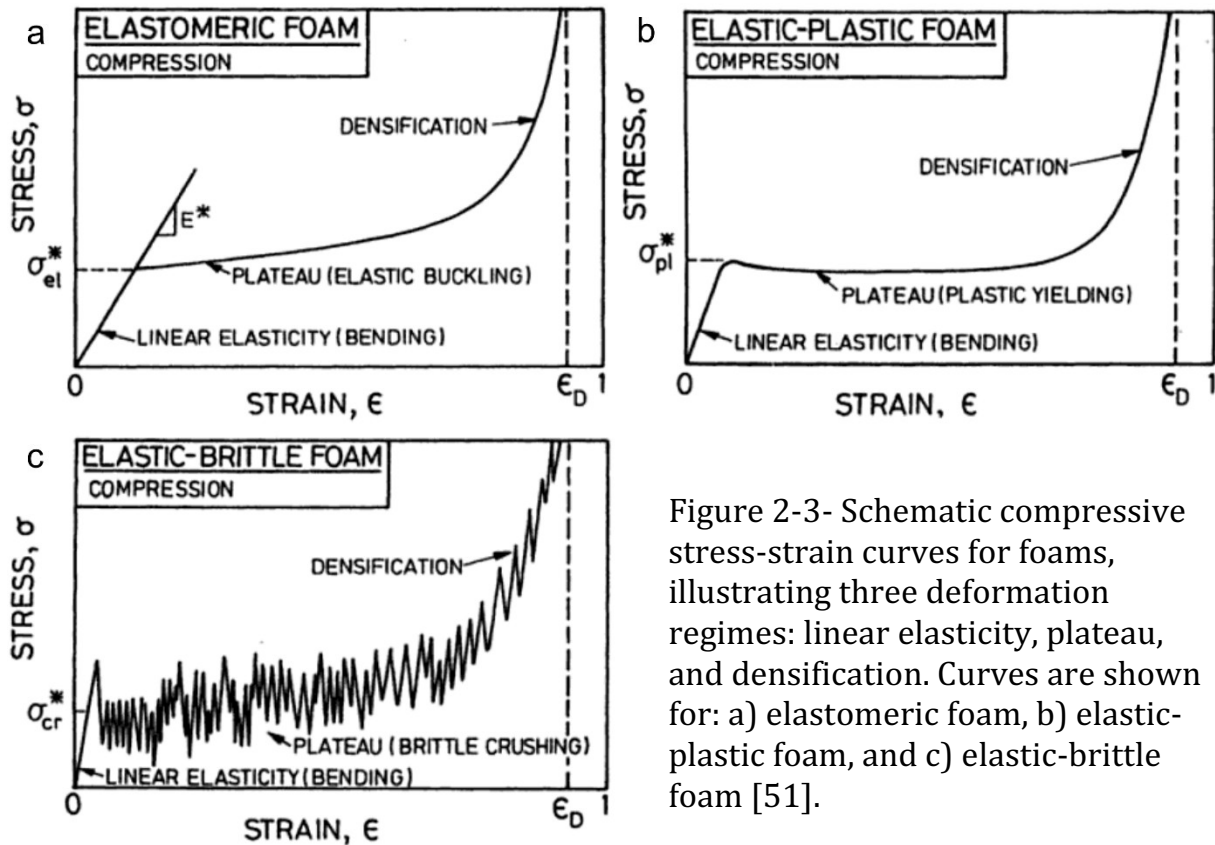


Figure 2-3- Schematic compressive stress-strain curves for foams, illustrating three deformation regimes: linear elasticity, plateau, and densification. Curves are shown for: a) elastomeric foam, b) elastic-plastic foam, and c) elastic-brittle foam [51].

Plateau region: The plateau deformation of foams begins with the plastic deformation of the struts or walls. During this stage, the cells begin to collapse plastically by the formation of plastic hinges in a foam that yields (Figure 2-3b), or by brittle crushing in a brittle foam (Figure 2-3c). The schematic deformation of cells for different base materials according to Figure 2-3 (elastomeric, ductile, and brittle) for an idealized cell in an open-cell foam designed by M.F. Ashby [16] is shown in Figure 2-4.

Once the stress applied to the foam exceeds its elastic limit, individual cells start to deform plastically by forming plastic joints at corners (Figure 2-4c), or they collapse sequentially as the cell edges break (Figure 2-4d). As some cells collapse, the load is redistributed to neighboring yet intact cells, allowing the overall stress to remain stable. This progressive collapse continues throughout the material, providing a consistent level of stress across a range of strain. The plateau region is particularly valuable in applications requiring consistent energy absorption during deformation, such as crash protection systems and cushioning materials.

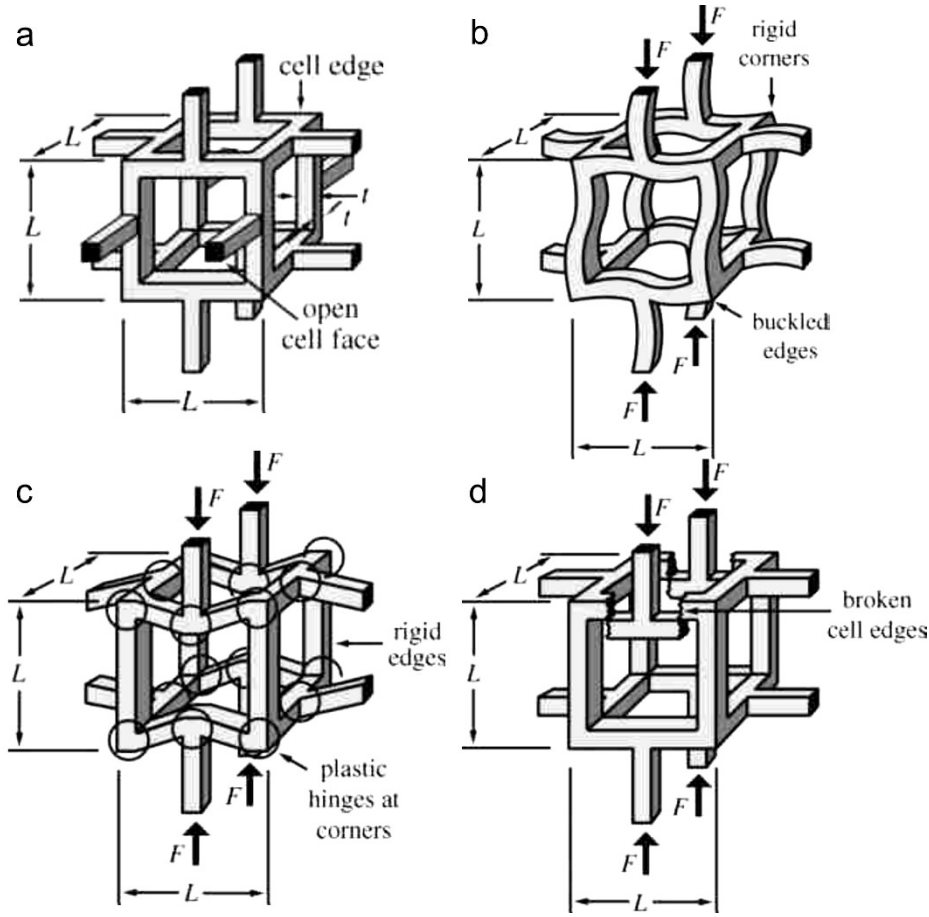


Figure 2-4- An idealized cell in an open-cell foam, designed by M.F. Ashby a) without deformation, made of b) elastomeric, c) ductile, and d) brittle materials [16].

The plateau stress (σ_{pl}) of foams can often be expressed as Equation 2 [5,9,51].

$$\frac{\sigma_{pl}}{\sigma_{ys}} = k \cdot \rho_r^n \quad \text{Equation 2}$$

where k is another material constant, σ_{ys} is the yield strength of the material, ρ_r is the relative density, and n is typically around 1.5 for open-cell foams and close to 2 for closed-cell foams. This relationship indicates that compressive strength increases with increasing relative density, though the exact exponent may vary based on the material and its structure.

Any parameter that changes the deformation behavior from purely ductile to higher brittleness causes fluctuations in the stress-strain curve, as seen in Figure 2-3c. This includes the alloying of pure metals, such as Al-Si alloys, which introduce brittle phases into the microstructure. In pure Al foams, both open- and closed-cell types, structural collapse occurs primarily through the buckling of the structural elements across the entire deformation range. This buckling is typically evenly distributed throughout the sample, leading to smooth stress-strain curves [54,74]. In contrast, Al alloy foams collapse by fracture of the struts/walls, and fewer buckling modes

are observed [54]. The sudden collapse of cells results in a reduction of stress, which is reflected as fluctuations in the stress curve.

Densification region: Once most of the cells have collapsed and the material is tightly packed, it enters the densification region. Here, stress increases sharply as the remaining voids are eliminated, and the material begins to behave more like a solid [5,9,70]. The onset of densification, referred to as the densification strain, is a critical parameter for characterizing the mechanical properties of cellular solids. For example, understanding the densification strain is essential when calculating the energy absorption of a cellular solid. The critical strain at the start of densification correlates with the initiation of interactions between cell walls, where contact between cell walls suppresses the usual deformation modes that characterize the plateau region [75].

There are several methods to determine the densification strain. It can be defined in the following ways: 1) by the intersection of the tangents to the stress plateau and densification regimes, 2) as the strain at the last local minimum before the stress increases sharply, and 3) as the strain at which the slope of the tangent matches that of the elastic regime [75]. However, in ductile foams, such as pure aluminum or austenitic stainless steel, the transition points are relatively smooth, making them difficult to determine reproducibly using these common methods. It has been found that investigating the strain hardening behavior may provide a valid approach for calculating the transitions between the characteristic regions of ductile metal foams [47].

Strain hardening: In ductile foams, no distinct plateau stress is observed. Instead, the foam undergoes continuous hardening throughout the entire strain range. The increase in stress during plastic deformation in metal foams can be attributed to strain hardening within the material itself, followed by further hardening at a later stage due to the interaction of struts or walls with each other [11,47,74]. This behavior can be expressed as Equation 3 [47].

$$\frac{d\sigma}{d\varepsilon} = \gamma + a \cdot \varepsilon^b \quad \text{Equation 3}$$

where γ is the linear strain hardening coefficient of the base material and $a \cdot \varepsilon^b$ is the hardening term caused by densification.

As in bulk materials, strain hardening in metal foams is influenced by the microstructure of the foam struts, to some extent independent of the relative density [76]. Additionally, a crucial effect arises from the re-orientation of the struts during loading. The densification strain can be defined as the point where the

hardening induced by the structural interaction of the struts surpasses the linear strain hardening of the base material ($a \cdot \epsilon^b > \gamma$) [47]. This occurs when the stress derivation in the plateau region exceeds the strain hardening coefficient. A similar approach can be applied to determine the transition from the elastic to the plateau region for ductile foams that exhibit a smooth transition between these regions.

2.4.3. Energy absorption

The energy absorption per volume of the metal foam (W) is defined as the area under the stress strain curve up to the densification strain (ϵ_0) as Equation 4 [47,57,77].

$$W \text{ (MJ/m}^3\text{)} = \frac{1}{100} \int_0^{\epsilon_0(\%)} \sigma \text{ (MPa)} \cdot d\epsilon \quad \text{Equation 4}$$

As can be interpreted from the formula, the energy absorption of metal foams is linked to the plateau stress and densification strain. Therefore, it can be enhanced by modifying the foam topology (mesostructure), increasing the relative density, or improving the strength of the strut material. Typically, open-cell foams, which undergo large strains at nearly constant stress, allow for significant energy absorption.

The energy absorption efficiency (Φ) is defined as the actual energy absorbed by the metal foam divided by the energy absorbed by an ideal absorber [47,57]. An ideal absorber reaches the maximum stress directly and maintains it constant throughout the entire deformation process [51]. In metal foams, it can be calculated using Equation 5.

$$\Phi = \frac{W}{\sigma_{\max} \cdot \epsilon_0} \quad \text{Equation 5}$$

where $\sigma_{\max}$ is the maximum stress up to the actual deformation and ϵ_0 is either the densification strain or the strain at the actual deformation point. A perfect energy absorber has an energy absorption efficiency of 1, while an elastic-brittle material has an efficiency of 0.5 [51].

2.4.4. Bending- and stretch-dominated structure

For cellular solids, understanding the balance between bending- and stretch-dominated structures is crucial to explaining their response to mechanical loading and how their architectural design influences performance. The straightforward rules for predicting the mechanical behavior of cellular materials based on relative density, such as those presented in Equations 1 and 2, may also be modified when the joining of elements leads to a different structural architecture.

Bending-dominated structure: This refers to a structural configuration where the primary response to applied loads involves rotation of struts about the joints when the joints are not locked, and bending of the struts when the joints are locked, rather than direct axial stretching. The deformation mode is primarily influenced by the geometry and connectivity of the cells in the structure. Bending-dominated structures or mechanisms are typically found in foams or cellular materials with low connectivity between cells, where the individual components are not part of a highly rigid network [16]. Examples of mechanisms can be found in Figure 2-5. These materials generally exhibit lower stiffness and lower initial yield strength compared to stretch-dominated structures.

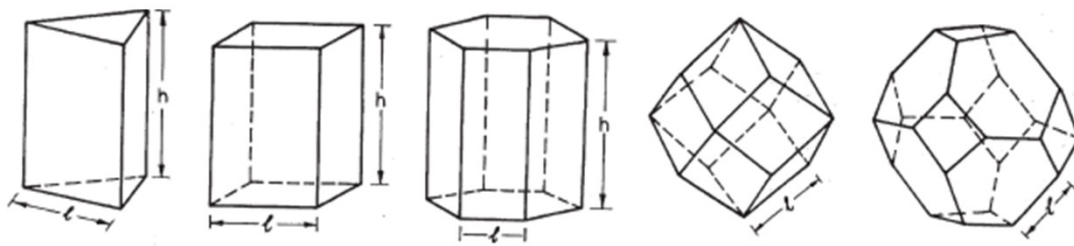


Figure 2-5- Polyhedral cells with low connectivity showing examples of bending-dominated structures [16].

The bending of cell walls or struts allows for progressive deformation, effectively dissipating energy. As the walls bend, they do not fail instantly but instead deform plastically over a range of strains. This characteristic results in a prominent plateau region on the stress-strain curves of bending-dominated foams, where the stress remains relatively constant as the cells collapse. This behavior is the main reason that makes bending-dominated structures particularly well-suited for energy-absorbing applications.

Stretch-dominated structure: In contrast, in a stretch-dominated structure, the primary response to applied loads is the axial expansion or compression of the cell walls or struts, rather than bending [22]. This type of deformation arises from specific architectural arrangements, typically characterized by high connectivity and rigidity in the cell structure. The struts in these structures are arranged to bear loads axially, resulting in higher stiffness and strength per unit mass compared to bending-dominated structures [16]. Examples of stretch dominated structures can be observed in Figure 2-6.

Stretch-dominated configurations are preferred for structural applications where load-bearing capacity and high strength-to-weight ratios are crucial, as they offer more efficient load transfer with minimal material deformation [22]. A stretch-dominated foam with a relative density of 0.1 is expected to be approximately ten

times stiffer and about three times stronger than a bending-dominated foam with the same relative density [78]. These structures are less susceptible to large deformations under load and can maintain their shape and rigidity.

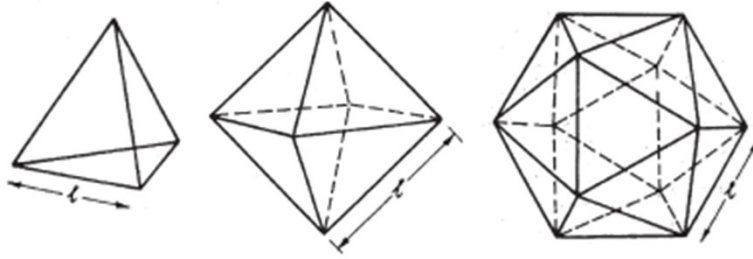


Figure 2-6- Polyhedral cells with high connectivity serve as examples of stretch-dominated structures [16].

Maxwell's stability criterion: The Maxwell criterion states that the stability of a structure can be assessed by comparing the number of constraints (connections) to the degrees of freedom within the structure. By examining these connectivity requirements, we can determine whether a structure is bending-dominated or stretch-dominated.

The condition for a pin-jointed frame to become both statically and kinematically rigid during loading in two dimensions is as Equation 6 and in three dimensions is as Equation 7 [16].

$$M = b - 2j + 3 = 0 \quad \text{Equation 6}$$

$$M = b - 3j + 6 = 0 \quad \text{Equation 7}$$

where M is the mobility or degree of freedom of the structure, b is the number of struts, and j is the number of joints or nodes. If $M < 0$, the frame is a mechanism and has one or more degrees of freedom. In the directions where displacement is possible, there are no load-bearing elements and the structure lacks stiffness [22]. If the joints are locked, the struts of the frame will bend when the structure is loaded [16]. If $M \geq 0$, the frame is no longer a mechanism. When loaded, its members carry axial loads, some in tension and others in compression, and it becomes a stretch-dominated structure [22]. Locking the joints now makes little difference, as slender structures are much stiffer when stretched than when bent.

The mentioned criteria are necessary conditions for rigidity, but not in general sufficient conditions as they do not account for the possibility of states of self-stress (a self-equilibrated state of strut tensions in the absence of external load) and of mechanisms. A generalization of the Maxwell rule in 3D is given by Equation 8.

$$M = b - 3j + 6 = s - m \quad \text{Equation 8}$$

where s and m are the numbers of self-stress and mechanisms respectively.

It should be noted that the performance of cellular materials is intrinsically linked to the base material, as it governs key properties such as strength, stiffness, and ductility. While metal foams can be produced from a variety of alloys, aluminum has proven to be particularly advantageous [37]. The unique properties of metal foams, combined with the low density, ductility, and strength of Al-based materials, result in outstanding overall mechanical performance [79]. Moreover, the wide range of aluminum alloys bring distinct advantages, enabling their adaptation to specific applications. Their ability to balance lightweight design with mechanical efficiency makes aluminum alloys an ideal candidate for the development of high-performance cellular materials. This sets the basis for more in-depth research into aluminum alloys for the end use of cellular structures.

2.5. Aluminum alloys

Aluminum, with an atomic number of 13 and an atomic weight of 26.98 g/mol, is one of the most abundant elements in the Earth's crust, accounting for approximately 8% by weight [80]. It is a lightweight metal with a density of 2.7 g/cm³, making it about one-third as dense as steel. In its pure form, aluminum has a tensile strength of about 90 MPa and a relatively low yield strength of around 35 MPa [80]. Its melting point is 660 °C, though this can decrease with alloying, enabling a wide range of processing techniques such as casting, forging, and additive manufacturing. These inherent properties make aluminum a versatile and effective base material for the development of high-performance alloys in cellular structures.

2.5.1. Aluminum-silicon alloys

Aluminum-silicon (Al-Si) alloys are widely used in the casting industry due to their superior castability, enhanced strength, and good wear properties. These alloys are characterized by their ability to form complex shapes with high precision. Alloying aluminum with silicon significantly lowers the melting point and narrows the solidification range, facilitating the casting process. The Al-Si system is a simple binary eutectic, where the key invariant transformation, aside from the melting of pure aluminum and pure silicon, is the eutectic reaction, in which the liquid transforms into α -Al (a solid solution with limited silicon solubility) and nearly pure silicon. This eutectic transformation occurs at 577 °C and 12.6 wt% silicon [80–82]. A thorough understanding of the nucleation and growth mechanisms of eutectic silicon particles is essential for controlling the final microstructure and morphology of the Si phase.

Microstructure evolution during the solidification of commercial hypoeutectic aluminum–silicon alloys occurs in two main stages [83]. The first stage involves the nucleation of primary aluminum grains at the liquidus temperature, followed by their growth in a dendritic form. The second stage comprises the nucleation and

growth of the aluminum-silicon eutectic. In commercial alloys, trace amounts of iron and phosphorus are typically unavoidable, resulting in the formation of intermetallic iron phases and aluminum phosphide (AlP), both of which play critical roles in the nucleation of eutectic phases [81–85]. Producing alloys from high-purity constituents can minimize these unwanted elements and alter the nucleation mechanism of silicon particles. Iron-containing intermetallics nucleate within the solute field ahead of the growing aluminum dendrites, often at or above the eutectic temperature of the alloy. Meanwhile, pre-existing AlP particles are pushed into the remaining liquid ahead of the dendritic front. Eutectic silicon then nucleates heterogeneously on these particles, followed by the nucleation of eutectic aluminum on the silicon [82]. The growth of the primary aluminum dendrites is arrested when the dendrites impinge on the eutectic Al grains afterwards.

It is notable that the dendritic growth of aluminum grains, as well as the size of the dendrites, affects both the distribution of heterogeneous nuclei and the solute concentration near the dendrite–liquid interface. These factors, in turn, influence the nucleation of silicon, which is essential for the cooperative growth of the aluminum–silicon eutectic. An increase in cooling rate reduces all kinds of dendrite arm spacings (DAS) [86–88], leading to a more uniform distribution of the eutectic phase. Additionally, increasing the silicon content also contributes to a reduction in DAS at a constant cooling rate [86].

2.5.2. Modification of eutectic silicon phase

Modification of the silicon phase in Al-Si hypoeutectic alloys is a critical step in improving their overall mechanical properties, particularly ductility and toughness [89–92]. This process primarily aims to transform the morphology of the eutectic silicon phase from coarse, brittle, plate-like structures to finer, more rounded fibrous forms [84,85,93]. On conventionally polished surfaces, these fibers appear as small individual particles; however, etching away the aluminum matrix reveals a fine coral-like structure as in Figure 2-7 for Al-8 wt% Si modified by 400 ppm Sr, 1 wt% Ce, and combination of both [94]. The flake morphology of silicon can be altered through three main approaches: increasing the cooling rate during solidification, reducing the concentration of natural inoculants present in the melt, and adding specific modifying elements [84,93,95].

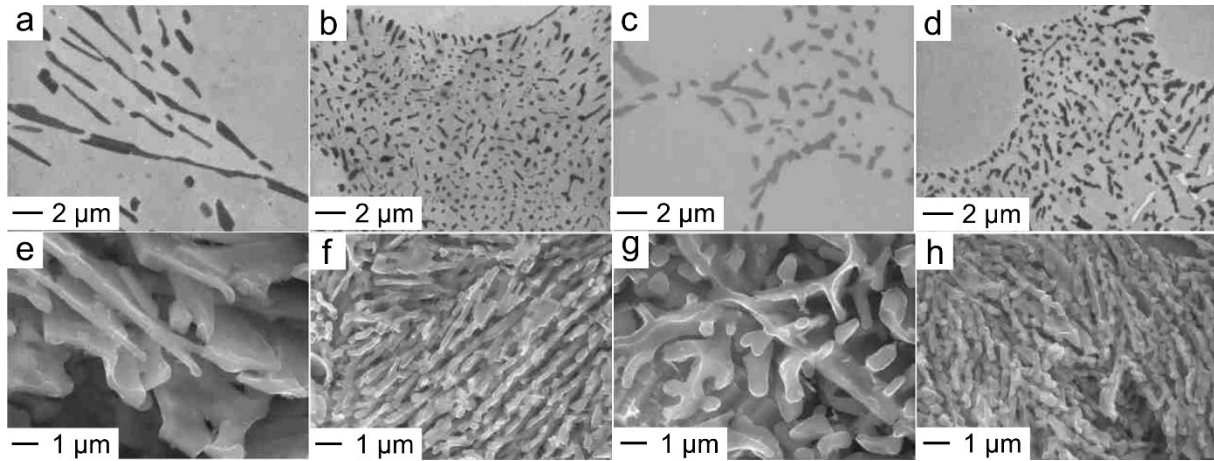


Figure 2-7- SEM images on unetched (a-d) and etched (e-h) samples of: a and e) Al-8Si, b and f) Al-8Si-0.04Sr, c and g) Al-8Si-1Ce, and d and h) Al-8Si-0.04Sr-1Ce [94].

The temperature at which the silicon phase begins to nucleate plays a key role in determining its final morphology. Rapid cooling delays the onset of nucleation and increases the degree of undercooling, defined as the difference between the actual temperature and the equilibrium solidification temperature. This results in a higher number of nucleation sites, leading to a finer silicon structure [93,96]. Additionally, higher cooling rates reduce the time available for silicon to grow into coarse, plate-like morphologies. As a result, increased cooling rates produce finer silicon particles and an overall refined microstructure. However, achieving uniform cooling throughout a casting with varying section thicknesses is often impractical in industrial settings.

To reduce the presence of natural inoculants in the melt, high-purity raw materials can be used. This approach lowers the phosphorus and iron content, thereby limiting the formation of unwanted and additional nucleation particles [84]. Another method involves increasing the melt temperature to fully dissolve any residual solid particles, further eliminating potential inoculants. However, both strategies are generally impractical for large-scale industrial applications due to their high cost and complexity. As a result, a more universally applicable and cost-effective method is required to achieve consistent microstructural modification.

Modifying elements alter the morphology of silicon by influencing both its nucleation and growth mechanisms. In non-modified commercial alloys, phosphorus is almost always present in the form of AlP, which serves as an effective nucleation site for silicon particles. This principle also underlies the refinement of hypereutectic alloys through phosphorus addition. However, when modifying elements are introduced, they can neutralize AlP particles by forming other compounds, causing silicon to nucleate at greater undercooling and thus leading to

finer structures [84,85,92]. The presence of chemical modifiers can also inhibit nucleation on intermetallic iron phases. During the final stages of solidification in the mushy zone, the concentration of modifiers in the eutectic liquid—especially within interdendritic regions—becomes relatively high. This increase raises the melt's viscosity and surface tension, which in turn elevates the wetting angle on intermetallic particles [81]. As a result, a large number of equiaxed eutectic aluminum grains nucleate before the eutectic silicon, forcing the silicon to grow between these grains and adopt a fibrous, broom-like morphology. The delay in eutectic nucleation is clearly indicated by the pronounced undercooling observed during thermal analysis of the solidification process [83,96–98].

The most important chemical modifiers with industrial application for retarding eutectic nucleation are sodium (Na) and strontium (Sr) [81,83]. These elements deactivate phosphorus-based inoculants that are active in the melt [96]. Although over-modification by sodium is more critical, it is more difficult to control the exact content in the melt due to high fading by evaporation or oxidation and low absorption efficiency [85,89,93]. Strontium, by contrast, has a high recovery rate and provides a more stable, long-lasting modifying effect [85,89]. For a modifier to be effective, it must not only limit nucleation but also alter the growth kinetics of silicon crystallizing from the melt, ultimately leading to the formation of a fibrous eutectic structure.

It is believed that the growth of coarse, faceted silicon particles occurs through the attachment of silicon atoms to twin plane re-entrant edges (TPRE), leading to the formation of long, thin particles with their longitudinal axis aligned along the twin planes [84]. Effective modifiers influence the twin density in the silicon phase either by poisoning the TPRE sites or by promoting twin formation. This modification mechanism has been investigated for various elements [93,97,99]. Elements that promote twinning tend to have larger atomic radii than silicon, with effective modifiers typically falling within a radius ratio range of 1.54 to 1.85 [93]. More detailed studies on rare earth elements and lanthanides have shown that a radius ratio of approximately 1.65 is the most effective [92,99,100]. The atomic radius ratios for the lanthanides, Na, and Sr relative to the atomic radius of Si are shown in Figure 2-8. Among the elements tested, apart from Na and Sr, europium was the only one shown to produce fully modified, fine-fibered silicon [100–102].

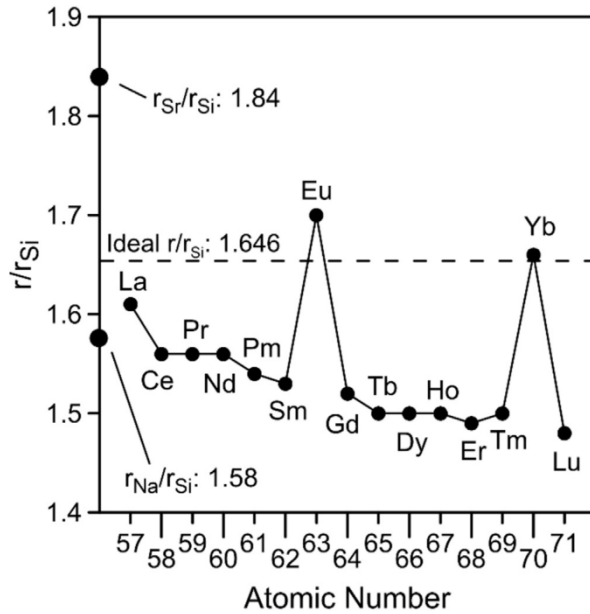


Figure 2-8- Ratio of atomic radii (r) for lanthanides, sodium, and strontium relative to the atomic radius of silicon (r_{Si}). The ideal ratio line at 1.646 identifies europium as a perfect match [99].

Microalloying with strontium (Sr) is considered one of the most practical methods for modifying the silicon phase. Sr co-segregates with aluminum (Al) and silicon (Si), forming nanometric segregations that promote impurity-induced twinning and restrict the growth of the eutectic Si phase [103]. This modification results in a significant improvement in ductility, with an increase of over 100%, and a moderate enhancement in strength (up to 20%) in bulk alloys with Sr addition [90,104]. However, these benefits are diminished by a reduction in the cooling rate during solidification [105,106].

Despite these advantages, modifying elements can also introduce negative side effects, such as increased porosity, hot cracking, and poor surface quality [92,104]. Additionally, the changes in nucleation patterns caused by modification can lead to complications [92]. Specifically, the crystallographic orientation relationships between the aluminum and silicon phases that are present in the unmodified state no longer apply after modification [107].

2.5.3. Iron intermetallic phases

The energy required to produce secondary or recycled aluminum is only about 5% of that needed for primary aluminum production [108,109], making secondary aluminum production highly cost-effective. However, the accumulation of impurities such as iron (Fe), magnesium (Mg), copper (Cu), and zinc (Zn), which are difficult to remove during recycling, poses a significant challenge [110]. Furthermore, the solubility of impurities like iron is high in molten aluminum but very low in solid aluminum, resulting in the precipitation of these impurities during solidification [111].

Iron is one of the most detrimental impurities in cast aluminum alloys, as it leads to the formation of brittle intermetallic compounds [108,112]. Some of these intermetallic particles solidify before the primary α -Aluminum grains, referred to as primary particles, and adversely affect the casting process [109]. Depending on the chemical composition and casting conditions, iron intermetallics can form in various structures, such as α - $\text{Al}_8\text{Fe}_2\text{Si}_3$ [108,113], α - $\text{Al}_{12-16.5}(\text{Fe,Mn,Cr,Cu})_{2.3-3.3}\text{Si}_2$ [114], β - Al_5FeSi [108,113] or π - $\text{Al}_8\text{FeMg}_3\text{Si}_2$ [108,113]. The size, morphology, and volume fraction of these Fe-containing intermetallics significantly impact the mechanical properties of Al-Si alloys. Among these, the β -phase is primarily responsible for the deterioration of mechanical properties, as it serves as a potential site for crack initiation [108,110–115]. This phase typically forms thin plate-like structures that appear as needle-like shapes in cross-sections (Figure 2-9a). These platelets are hard and brittle, with relatively weak bonding to the matrix [110,112]. As the iron content of the alloy increases, the number of Al_5FeSi crystals does not significantly increase, but they tend to grow into large platelets [112].

The size of β - Al_5FeSi crystals and the likelihood of their formation increase not only with higher iron content but also with decreasing cooling rates [112,115]. To control their size and minimize their detrimental effects, common methods include rapid solidification at high cooling rates, which lowers the formation temperature of the β -phase below the silicon-aluminum eutectic temperature, thus promoting the formation of the α - $\text{Al}_8\text{Fe}_2\text{Si}_3$ phase. Alternatively, adding manganese (Mn) encourages the formation of the α - $\text{Al}_{15}(\text{Fe,Mn})_3\text{Si}_2$ phase as can be observed from Figure 2-9 [108,110]. Additionally, it has been shown that the thermal history of the molten alloy, including factors like superheating and holding time, can alter the solidification path [110]. Both α - $\text{Al}_8\text{Fe}_2\text{Si}_3$ and α - $\text{Al}_{15}(\text{Fe,Mn})_3\text{Si}_2$ phases can form in a “Chinese-script” morphology or as compact globular particles, both of them are less harmful to the mechanical properties of the alloy.

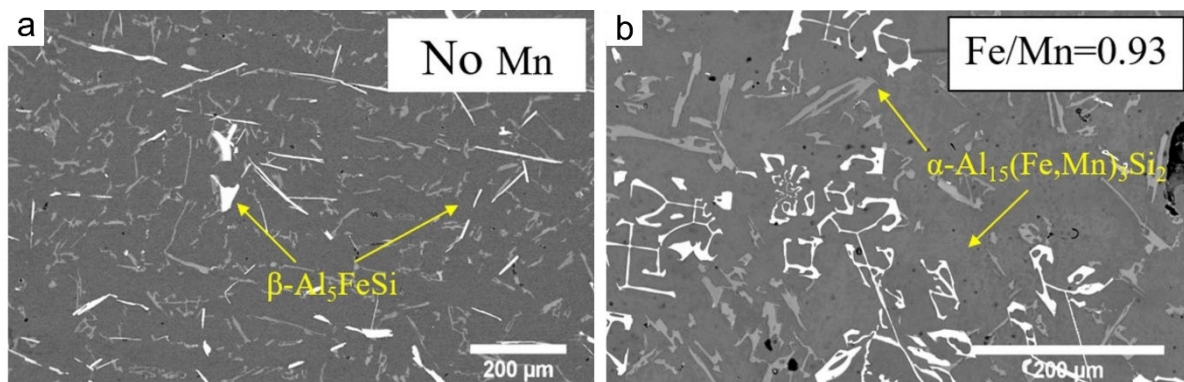


Figure 2-9- Back-scattered SEM images of Al8SiMg alloy containing 0.7wt% Fe a) without Mn and b) with 0.75 wt% Mn addition [108].

The standard approach for minimizing the negative effects of iron on the mechanical properties of cast aluminum alloys is to maintain a manganese concentration that is at least half the iron content [108,109]. However, the critical Mn content required to eliminate the β -Al₅FeSi phase is not only dependent on the alloy composition but also on the cooling rate [109]. At lower cooling rates, higher Mn concentrations are needed to effectively suppress the formation of the β -phase [110].

2.5.4. Aluminum-copper alloys

Aluminum-copper (Al-Cu) alloys are a significant class of materials, valued for their outstanding mechanical properties and adoptability through heat treatment [116]. These alloys are the scope of interest from the early work of Guinier and Preston since 1938 [117] and primarily derive their strength from solid solution strengthening and precipitation hardening [118]. The maximum solid solubility of Cu in Al is 5.65 wt% at the eutectic temperature of 548.2 °C and it decreases when temperature drops [119]. Precipitation hardening, through a carefully controlled heat treatment process, enables the formation of copper-rich precipitates within the aluminum matrix, enhancing the alloy's strength.

The heat treatment process for Al-Cu alloys begins with a dissolution stage, occurring within the temperature range of the single-phase aluminum solid solution. During this stage, all copper-containing precipitates from solidification dissolve into the aluminum matrix [116]. This is followed by a quenching process that retains all copper atoms within the matrix. Nucleation and growth of the Al₂Cu precipitate occur as the supersaturated solid solution decomposes during the subsequent aging process [116]. The precipitation sequence in Al-Cu alloys typically progresses through the formation of Guinier-Preston (GP I) zones, followed by the formation of θ'' , θ' (intermediate phases), and ultimately the equilibrium θ (Al₂Cu) phase [118,119].

The effects of different precipitates on the mechanical properties of aluminum-copper alloys, namely strength, hardness, and ductility, are closely related to their size, distribution, coherency with the aluminum matrix, and the stages of the precipitation process [118]. GP zones are the smallest and most finely dispersed precipitates, remaining coherent with the aluminum matrix (Figure 2-10a). These clusters of copper atoms distort the matrix lattice due to their size mismatch with aluminum, significantly contributing to initial strengthening through lattice distortion, which hinders dislocation motion. The small size and high dispersion of GP zones maintain good ductility while enhancing strength. The θ'' phase is a metastable phase that also remains coherent with the matrix (Figure 2-10b). It consists of thin, plate-like structures that cause greater lattice distortion than GP

zones. θ'' provides peak strengthening by strongly hindering dislocation movement while maintaining coherency with the matrix (Figure 2-10e). However, ductility starts to decrease compared to the GP zone stage as the precipitates grow, and stress concentrations increase. The θ' phase, a larger metastable phase, is semi-coherent with the matrix and forms with further aging at higher temperatures or longer durations (Figure 2-10c). θ' offers lower strength than θ'' due to its semi-coherency, weakening the dislocation hindrance effect. Ductility is further reduced as the precipitate size increases, making them act as stress concentrators and promoting microcrack formation. The θ phase, the equilibrium phase, is completely incoherent with the aluminum matrix (Figure 2-10d). These larger, coarsely distributed precipitates significantly reduce strength, as incoherent precipitates offer minimum resistance to dislocation motion, and coarsening reduces the total number of strengthening particles. The aging curves of Al-Cu alloys with various Cu concentrations are shown in Figure 2-10e and f, where the Al-Cu binary alloys are solution treated in the single α -phase region, water quenched to room temperature, and then aged at 130 °C or 190 °C. The relationship between hardness or strength and the type of precipitates is clearly visible in these graphs.

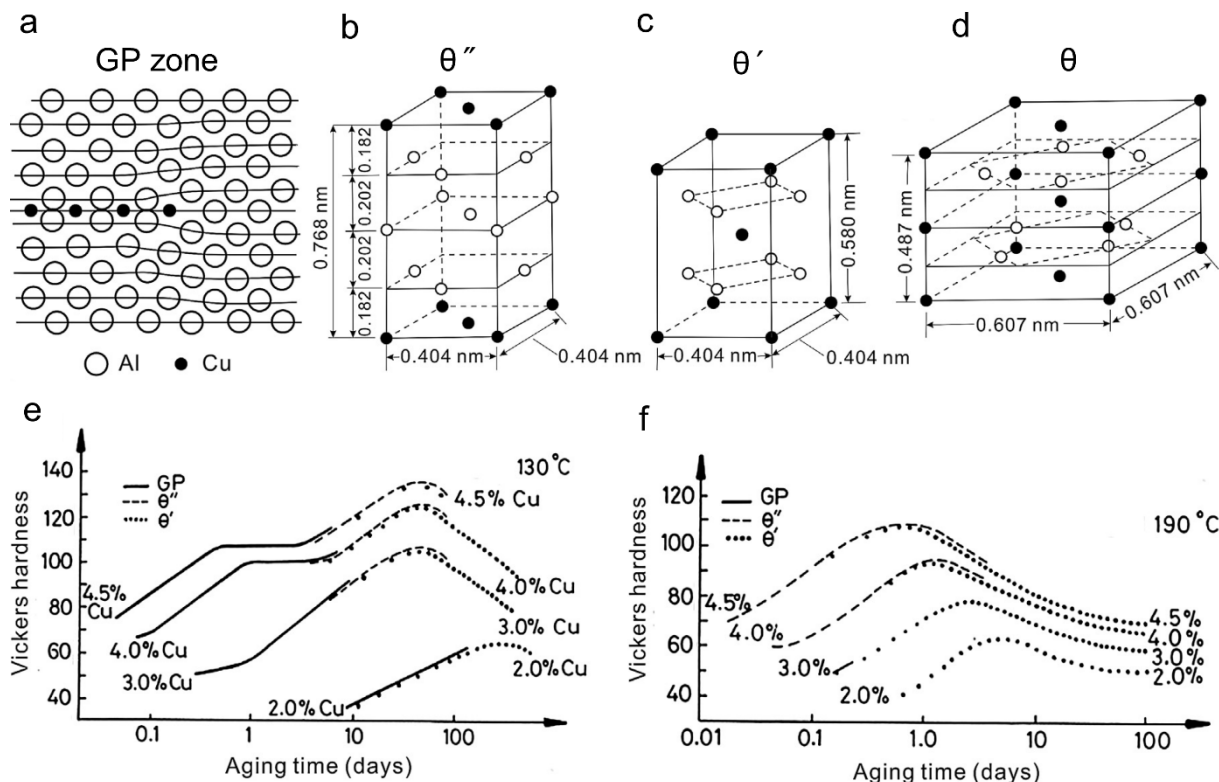


Figure 2-10- Various precipitates of the Al-Cu alloy system resulting from continuation of heat treatment: a) GP zone, b) θ'' , c) θ' , and d) θ . Effects of alloy composition and aging temperature on the maximum hardness achievable in Al-Cu binary alloys: e) by aging at 130 °C and f) by aging at 190 °C [119].

The balance between strength and ductility in aluminum alloys can be achieved by carefully controlling the aging temperature and time, in order to optimize the size and distribution of precipitates. The same principle can be applied to the cellular aluminum structures by using the same alloying concept. By adjusting the size of the precipitates through heat treatment and controlling the ratio of precipitate size to strut or wall thickness, the overall mechanical behavior of the cellular materials can be influenced.

2.6. Aluminum alloy cellular structures

Aluminum alloy cellular structures stand out among metallic cellular materials due to a unique set of characteristics that make them highly versatile and valuable across various applications. Aluminum is inherently lightweight, and this is amplified when it is structured in a cellular form. Aluminum cellular structures provide strength with minimal weight, which is advantageous in industries like aerospace, automotive, and construction, where reducing weight improves fuel efficiency, structural performance, and overall sustainability.

Aluminum cellular structures, particularly foams, are excellent at absorbing energy, making them effective in applications that require impact resistance and shock absorption. This quality is critical for crash protection components in transportation, where aluminum cellular structures can help dissipate energy during impacts and enhancing safety. Furthermore, aluminum has high thermal conductivity, a property that remains beneficial in its cellular form. Aluminum foams and lattices can dissipate heat efficiently while staying lightweight, making them suitable for heat exchangers, cooling systems, and other applications where temperature regulation is essential.

Aluminum cellular structures can be produced using various manufacturing methods, such as liquid metal foaming, powder metallurgy, casting routes, and additive manufacturing, each of which allows some degree of acceptable control over different structural levels. This flexibility enables engineers and designers to tailor aluminum cellular structures for specific applications and to balance trade-offs between cost, performance, and manufacturability.

2.6.1. Processing methods of aluminum cellular structures

Aluminum alloy foams can be produced using a range of techniques that differ in terms of the resulting cell type (open-cell or closed-cell), pore size, microstructure, and potential applications. When choosing the manufacturing process, a distinction is mainly made between the liquid processes and powder metallurgy. Different production routes are schematically illustrated in Figure 2-11.

Liquid metal foaming (direct gas injection or use of blowing agents): Liquid metal foaming, which includes gas injection or use of blowing agents, is a common method for producing closed-cell aluminum foams (Figure 2-11a). In the direct gas injection process, a gas (such as nitrogen, argon, or even air) is injected directly into a molten aluminum bath, creating bubbles that form a foam structure as the aluminum cools and solidifies [57,120]. To stabilize these bubbles during formation, solid particles like aluminum oxide, silicon carbide, or some elements like calcium are often added to the melt [70,121]. This addition serves to increase the viscosity of the molten metal, preventing the bubbles from merging and collapsing, which is crucial for maintaining a uniform cellular structure. Key parameters influencing this process include the gas flow rate, injection pressure, particle concentration, and cooling rate, all of which must be carefully controlled to produce foams with the desired cell size and distribution [122]. This technique is relatively simple and allows for continuous production, but the control over pore size and uniformity can be limited [123].

Alternatively, blowing agents can be mixed with molten aluminum, where they decompose at elevated temperatures to release gases like hydrogen. For instance, titanium hydride (TiH_2) is commonly used, decomposing to release hydrogen gas when heated [123–126]. This creates bubbles within the molten metal, which are trapped as the aluminum solidifies into a closed-cell foam; however as explained thickening agents like calcium or solid stabilizers are often needed to keep the foam structure stable during cooling. Parameters like the amount of blowing agent, temperature of decomposition, and melt viscosity play a critical role in determining the quality and uniformity of the foam [10]. These methods are capable of producing large volumes of foam blocks for different applications.

Powder metallurgy: The powder metallurgy method for producing closed-cell aluminum foams, often referred to as foaming of precursor materials (foamable precursors), involves mixing aluminum alloy powder with a foaming agent (Figure 2-11b) [127–129]. The foaming agent, the same as previous method, is selected for its ability to release gas (typically hydrogen) when heated. The aluminum powder and foaming agent are mixed and compacted into a dense precursor, often using methods like uniaxial pressing or extrusion, to form a solid billet or sheet. This precursor is then heated to a temperature above the melting point of aluminum alloy but below the decomposition point of the foaming agent. As the foaming agent decomposes, it releases gas, causing the molten aluminum to expand and form a closed-cell foam structure.

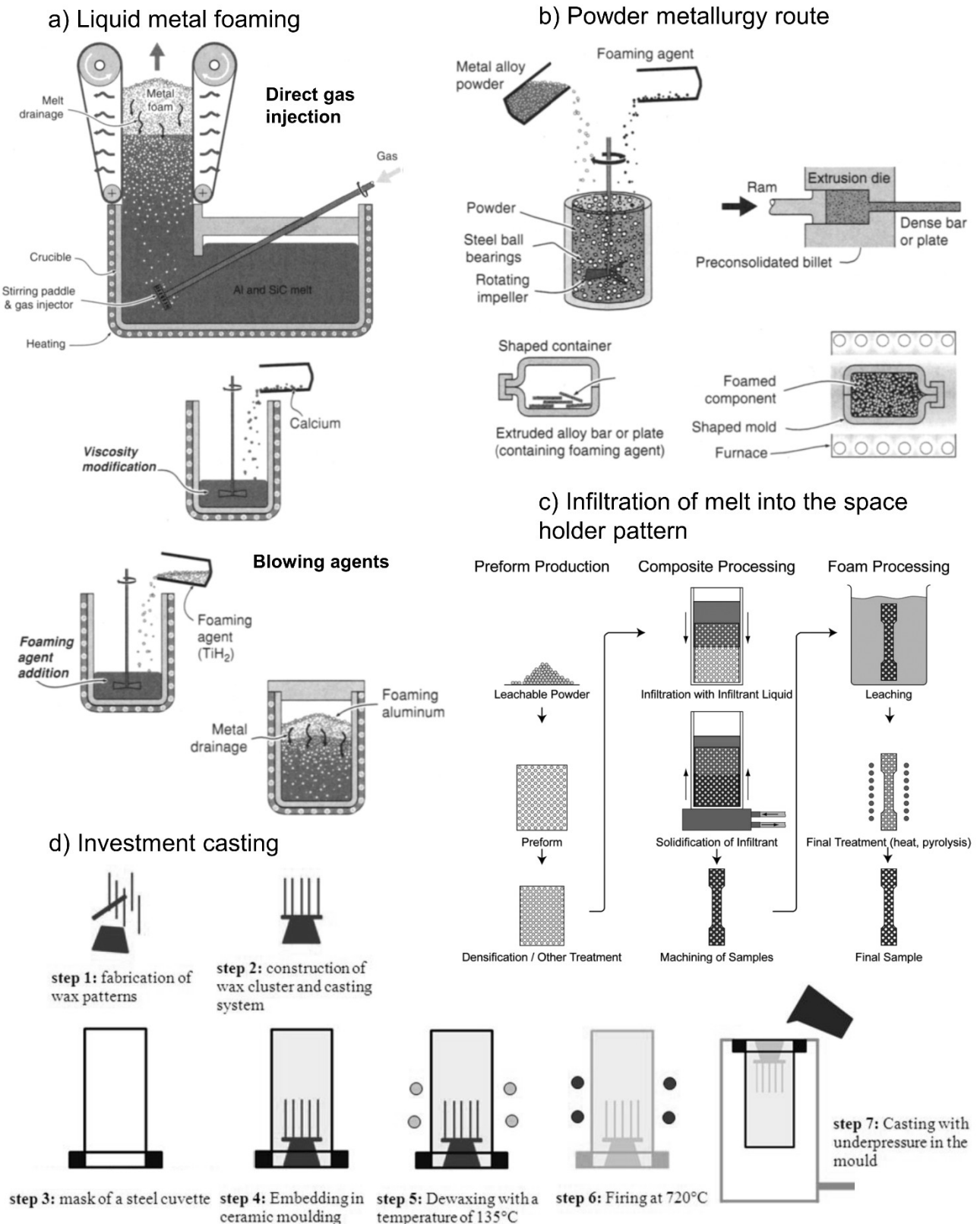


Figure 2-11- Processing of metallic lattice structures through a) liquid metal foaming with: direct gas injection and blowing agents, b) powder metallurgy route; casting routes: c) infiltration of melt into the space holder pattern and d) modified investment casting technique [70,130,131].

The process parameters, such as heating rate, temperature, and the amount of foaming agent, are critical for controlling the uniformity and size of the pores within the foam [129,132,133]. Higher heating rates can lead to rapid gas release and a more uneven foam, while precise temperature control helps achieve a stable structure with uniform cell size [132]. The compactness of the precursor also plays a role, as it affects the gas distribution and the ability of the metal to expand uniformly during foaming. This method allows for the production of closed-cell foams with controlled density and cell size particularly valued for its ability to produce complex shapes through pre-shaping of the precursors before the foaming process [132,133].

Space holder method: The space holder technique is a versatile method used for producing open-cell aluminum foams in both liquid route and powder metallurgy technique. In the powder metallurgy, the process begins by mixing aluminum powder with space holders, such as salt, carbamide (urea), or polymer beads, which act as temporary fillers [69,134]. This mixture is compacted into a desired shape and partially sintered to form a stable structure. The space holders maintain their form during this stage, creating voids in the metal matrix and are removed through leaching or dissolution after sintering, leaving interconnected pores for open-cell structures. If isolated voids are created through the process, the space holder material are entrapped in a closed-cell structure and cannot be removed.

Alternatively, the space holder technique can be used to create a pattern or a mold that is subsequently infiltrated with molten aluminum (Figure 2-11c) [11,37,135–137]. In this approach, the space holder mixture serves as a template placed inside a mold, into which the molten aluminum is introduced. The liquid metal fills the voids left by the space holders, and after solidification, the space holders are leached out, leaving a porous metal structure. This technique provides excellent control over pore size, shape, and connectivity, since from the beginning the structure geometry is defined by the shape of the space holders [137]. Adjustments to the infiltration process and the characteristics of the space holders allow for precise customization of the final foam properties.

Investment casting with sacrificial patterns: One of the common methods for producing open-cell aluminum foams is investment casting, using sacrificial patterns created through additive manufacturing or polymer-based templates (Figure 2-11d) [15,46,47,79,130,138,139]. This method involves employing a porous pattern, often made from wax, resin or polymer like polyurethane. The pattern can be shaped to match the desired geometry of the final foam component which is covered with a fine ceramic slurry out of investment powder to form a mold. For low melt

temperature alloys like aluminum, gypsum bonded investment powder and for high melt temperature alloys like steel, phosphate bonded investment materials are used as molding materials. Once the investment mold is formed and dried, the sacrificial pattern is melted or burned out, leaving a negative mold of the desired porous structure. Molten alloy is then poured into this mold, filling the voids and forming a network of interconnected pores. After the aluminum solidifies, the ceramic mold is washed away, leaving the open-cell aluminum foam.

In order to fill the mold cavity as expected, some precautions are necessary to apply. Due to the fine geometry of the foam structure and struts, pre-mature solidification can stop melt flow before complete mold filling. Therefore, usage of molds at high temperature is the main key aspect for the casting process, which will result in low cooling rate [46,71,79]. On the other hand, over pressure on the melt or under pressure on the perforated mold flask is necessary. The over pressure can also be understood in specific casting processes like centrifugal casting both for alloys with high and low melt temperatures [47,71,140]. The process also allows to some extent for a control over the foam's microstructure by adjusting parameters such as pouring temperature, filling speed, and cooling rates [71,79].

One of the advantages of using sacrificial patterns in investment casting is the ability to produce complex shapes and features that would be challenging to achieve with traditional production methods [38,41,43]. Furthermore, this method allows for precise control over the pore geometry and distribution based on the initial pattern [35]. The customized structural patterns can be produced through additive manufacturing (AM) and integrated into the investment casting process as a hybrid investment casting or indirect additive manufacturing technique.

Additive manufacturing for production of lattice structures: In order to benefit from the advantages of the fine-tuning of the geometry in aluminum lattice structures, sophisticated manufacturing methods should be employed. Recent improvements in additive manufacturing (AM) techniques, such as design flexibility and printing resolution, have revolutionized the fabrication of these complex structures [29–31]. AM enables the fabrication of complex and optimized lattice structures that are difficult to achieve through traditional manufacturing. The layer-by-layer nature of this process allows for precise control over lattice design, which can be customized to meet specific structural and functional requirements.

AM allows for fine-tuning the lattice parameters, such as strut diameter, unit cell geometry, and relative density, directly impacting the mechanical properties of the final structure. The precise control over design enables the creation of stretch-dominated or bending-dominated structures, depending on the application

requirements. AM is a near-net-shape process and a high material efficiency of it reduces waste, which is essential when working with higher-cost materials like specific aluminum alloys.

Two of the most commonly used direct AM production methods are Laser Powder Bed Fusion (LPBF) [30,32,141] and Selective Laser Melting (SLM) [142,143]. In both techniques, the laser power, the scanning strategy, the deposition parameters, and the hatch spacing have significant impacts on the geometry and quality of the struts and the lattice performance [26,32]. In these processes, various defects like internal voids, presence of un-melted powder on the surface of the struts, irregular cross-section, and missing struts can occur [144]. Furthermore, due to the different thermal conductivity of loose powder and solid part, the solidification and microstructure of inclined and vertical struts are different [60].

On the other hand, the indirect use of additive manufacturing for the production of lattice structures can be adopted as an attractive alternative with great potential. Most of the indirect additive manufacturing processes combine the 3D printing of patterns utilized in lost pattern casting techniques. The integration of additive manufacturing into casting makes it possible to benefit from the design freedom of AM and has offered several practical aspects for many years [38,41].

In particular, stereolithography (SL) [42,43] and digital light processing (DLP) [38] techniques have been employed to create detailed 3D-printed patterns from castable resins. These techniques use a resin that hardens when exposed to a laser source or UV light, enabling the production of highly accurate, three-dimensional objects with a fine resolution. Fused deposition modeling (FDM) [35,39,40,145] and binder jet methods [34] have also been effectively used for producing patterns intended for casting. In FDM, a filament is heated and melted to allow for layer-by-layer deposition, which can offer cost benefits over stereolithography but typically with a lower resolution.

Processing techniques play a crucial role in shaping the structure and properties of aluminum cellular materials, influencing everything from pore size to mechanical performance. However, the material's inherent characteristics, determined by its alloy composition, are equally important in defining the final performance of these structures. Based on the understanding of the processing methods, the following section explores the contribution of aluminum alloys to the development of high-performance cellular structure.

2.6.2. Pure aluminum foams

Aluminum in the pure form creates the most ductile condition for the aluminum cellular structures. In both open-cell and closed-cell foams, pure aluminum foams deform homogeneously under compression and the stress-strain curves show very smooth transitions from quasi elastic to plateau region and from plateau to densification region [11,20,37,74]. Due to the high ductility during deformation of struts or cell walls, high energy absorption can be expected from aluminum foams. However, a pronounced strain hardening response can be observed during deformation, which affects the plateau region and results in the stress-strain curves being inclined at this stage [74].

The most important factor influencing the mechanical properties and above all the energy absorption of pure aluminum foams is the relative density [20,74]. With the increase in relative density, the stress-strain curves show three obvious changes: the plateau stress increases (as suggested by the Gibson-Ashby power law, Equation 2), the length of plateau region becomes shorter (densification stage starts earlier), and the plateau region becomes more inclined, which indicates the strain hardening becomes more evident [20]. These changes can be observed for closed-cell pure aluminum foams as Figure 2-12.

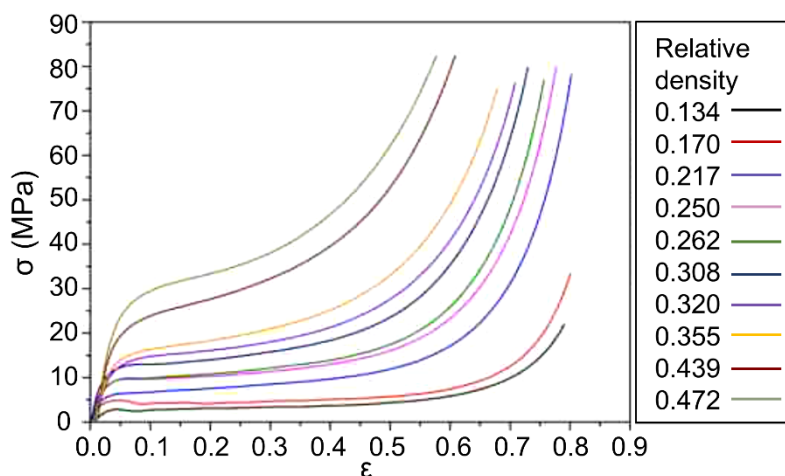


Figure 2-12- Quasi static compression test curves of closed-cell pure aluminum foams by changes in relative density [20].

Pure aluminum foams, while advantageous for excellent energy absorption suffer from significant mechanical limitations due to the inherent low strength of aluminum in its pure form. Alloying addresses these weaknesses by incorporating elements like silicon, magnesium, or copper, which enhance mechanical properties through mechanisms like solution strengthening, precipitation hardening, and grain refinement. For instance, aluminum alloys such as Al-Si or Al-Si-Mg exhibit considerably higher yield and tensile strengths, making them more suitable for structural applications. By taking advantage of the benefits of alloying, aluminum

foams can achieve a balance between low weight and superior mechanical performance, expanding their applicability.

2.6.3. Cast aluminum alloy foams

As one of the main production methods for open-cell aluminum alloy foams, investment casting has proven to be very flexible, not only in terms of geometry of the structure but also in terms of the versatility of the alloys. Different alloys can be used for the production of aluminum foams, like Al-Si or Al-Si-Mg [15,53,72,79] and Al-Zn alloys [71,140] and alloy adjustments can be easily implemented in a similar way to a traditional casting process.

The investment molds used for casting open-cell foams typically exhibit low thermal conductivity and high heat capacity [15,71,130,140]. To ensure proper mold filling and prevent premature solidification, precautions such as elevated mold temperatures are employed as well [139]. As a result, the cooling and solidification rates of the foams are relatively slow.

Previous researches have shown strong scattering in the micromechanical properties of cast foams. This scattering has been attributed to micro-porosity and inclusions from the casting process, as well as the irregular shape, rough surface, and microstructural inhomogeneities of the struts [53]. Therefore, it was suggested to conduct tests on individual struts rather than the whole foams to gain a more precise understanding of the mechanical behavior. Shrinkage porosities typically form at the nodes due to the high node-to-strut thickness ratio, which results in greater mass accumulation and localized non-uniform solidification [37]. However, casting defects within or near the nodes have not been observed to act as fracture initiation points, as the thin struts experience higher stress concentrations compared to the nodes.

Observations during tests of open-cell foams showed that the microstructure of the struts has a great influence on the fracture behavior of the foam [15]. The microstructure of cast foams differs fundamentally from that of bulk material, primarily due to two factors: the small cross-sectional dimensions of individual struts and the slow cooling rate of the investment mold [140]. This slow cooling promotes the formation of extremely coarse grains, sometimes reaching the diameter of the struts themselves [15,71]. As a result, many cast struts exhibit a near-single-crystal structure [72,79]. Additionally, the presence of coarse silicon and intermetallic phases, significantly influences the mechanical properties of the foam [15].

It was found that local variations in the cooling rate over the entire investment mold are only possible to a limited extent. For example, the mass accumulation of the melt in the pouring and feeding area reduces the cooling rate compared to the other end of the foam [37,140] or the increase in the metal content by thickening the struts leads to a higher cooling rate within the foam [46]. However, due to the specific solidification behavior, this small change influences the microstructure and leads to a microstructural gradient, i.e. coarser aluminum grains were observed in the areas closer to the pouring section [37,71].

Considering the effect of the metallic base part of foams, aluminum alloy foams typically exhibit a jagged plateau, in contrast to ductile materials, which exhibit a uniform stress plateau in the stress-strain curve. This behavior is due to the brittle failure of individual foam struts when the local stresses exceed the strength of the strut material. When compressed, the individual struts are not only loaded axially, but also experience significant torsional and bending stresses, which lead to a C- or S-shaped deformation depending on the location of the plastic joint formation [18]. An example of an S-shaped strut that is developed during the deformation process can be seen in Figure 2-13. Due to the presence of numerous struts with different orientations, the foam undergoes a highly complex deformation process, with continuous changes in local loading conditions. As a result, struts either deform plastically through buckling or fail abruptly by fracture, with brittle phases (Si, Mg₂Si or AlSiFe) accumulating near the plastic hinges [18].

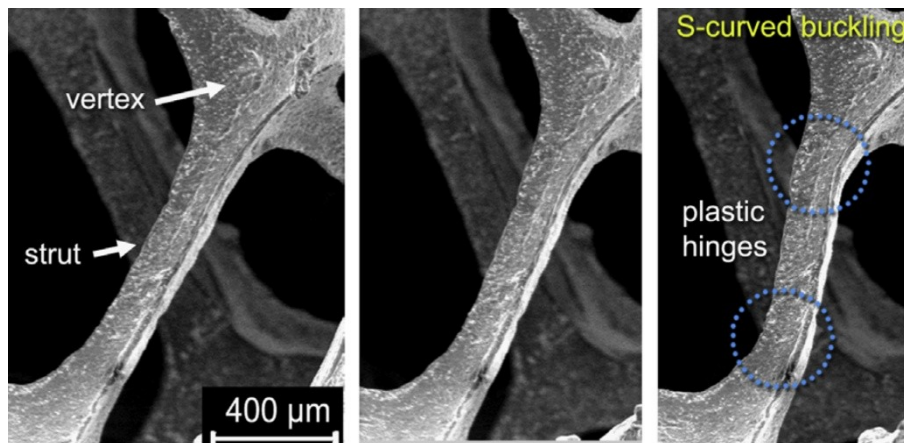


Figure 2-13- SEM micrographs showing the formation of the S-shaped strut during the compression test of an open-cell aluminum alloy foam [18].

While the mechanical properties of the base material can be significantly influenced by manufacturing parameters, direct correlations between bulk microstructural characteristics and foam microstructure remain limited. For example, in contrast to bulk aluminum alloys, where T6 heat treatment generally enhances strength, the strut strength in cast Al-Si-Mg foams was found to decrease after T6 treatment [72]. Consequently, the plateau stress of T6-treated foams remained nearly identical to that of as-cast foams. This can be attributed to the presence of coarse and angular

silicon particles, which do not readily undergo morphological changes during heat treatment, thereby neutralizing the strengthening effects that heat treatment would otherwise provide.

Moreover, several well-established principles for bulk cast materials do not directly apply to open-cell foams. For instance, in conventional castings, reducing melt or mold temperature typically results in finer dendrite arm spacing, reduced grain size, and an increased number of grains. However, in the case of cast foams, establishing a direct relationship between the casting conditions (e.g., mold temperature) and the resulting microstructure and thus the mechanical properties is far more complex due to the unique solidification dynamics within the thin struts [79].

Changing the geometry of the structural parameters, i.e. the strut length and cross-section as meso-structural parameters and the cell geometry as macrostructural parameter, can more effectively influence the mechanical properties of the material as a whole. However, the basic principles that influence the development of the microstructure remain the same within the same investment casting process.

2.6.3. Lattice structure versus foam

As previously introduced, metal cellular structures can be broadly classified into two types: stochastic porous foams and periodic cellular lattice structures. Stochastic foams feature a random distribution of open or closed voids, whereas periodic cellular lattices are based on repeating unit cells, resulting in a uniform structure [24,67]. These periodic structures can be designed with either a two-dimensional configuration, such as honeycombs, or a three-dimensional configuration, like strut-based lattice structures [29].

In stochastic foams, relative density is the dominant factor influencing mechanical properties, while cell size has only a negligible effect. However, cell shape plays a more significant role in determining foam strength [21]. In contrast, periodic lattice structures exhibit superior mechanical properties compared to stochastic foams at the same relative density [24,146]. This is due to the well-controlled geometry of lattice structures, which allows for precise tuning of mechanical behavior. In principle, a more predictable and controllable structure leads to more tailorable mechanical properties, making architected lattice structures highly versatile. Their mechanical performance can be adjusted over a broad range simply by modifying geometric design parameters [60].

Design principles and structural optimization: The design of lattice structures plays a crucial role in achieving desired mechanical properties. Early structural principles were inspired by nature, which has long served as an optimal model for cellular

solids [28]. Over time, engineering adaptations have introduced specific design strategies, including relative density control over the structure [31], the classification of structures as bending-dominated or stretch-dominated [22], and more recently, computational simulations and optimization techniques introduced further advanced lattice design, enabling structures to be tailored for specific mechanical responses.

As mentioned before, switching from a bending-dominated structure to a stretch-dominated structure fundamentally alters mechanical performance. Stretching-dominated lattices exhibit significantly higher stiffness and yield strength than their bending-dominated counterparts at the same relative density [28]. This makes them ideal for lightweight structural applications requiring high mechanical stability. However, for energy-absorbing applications, bending-dominated structures are often preferable because they provide a stress-strain response with a long, flat plateau, which is essential for impact resistance [22,29].

The ability to design lattice structures with customized properties has led to the development of several innovative configurations. Auxetic cellular structure, for example, exhibit a negative Poisson's ratio (ν), meaning they expand laterally when stretched and contract when compressed [147,148]. This unique behavior has drawn considerable interest due to its potential for impact mitigation, enhanced shear resistance, and tunable stiffness. Functionally graded lattices feature a spatially varying solid volume fraction, allowing material properties to change gradually across different regions. This enables optimization for multiple performance criteria within a single structure [23]. By tailoring the unit cell design, lattice structures can be made anisotropic, meaning their mechanical properties differ along different loading directions. This is particularly useful for applications where directional strength and stiffness are required [67].

In this context, truss-based lattice structures have been extensively studied, and libraries of predefined unit cells with specific mechanical properties have been developed (Figure 2-14a) [33]. These unit cells can be selected based on the expected stress and strain conditions in real-world applications [67,144]. The spatial configuration of these structures is relatively simple, which is composed of rigid joints and uniform or variable cross-section struts [142,144]. Additionally, their simple yet effective spatial configuration allows for integration into larger-scale hierarchical components, further expanding their applicability in engineering design [144].

Recently, focus shifted towards cellular structures with mathematically defined architectures such as triply periodic minimal surface (TPMS) based topologies

(Figure 2-14b) [29]. These structures are characterized by a zero mean curvature in hyperbolic space, minimizing stress concentrations around structural nodes [142]. As a result, TPMS-based lattices provide enhanced energy absorption and improved mechanical efficiency throughout the loading process [30,142].

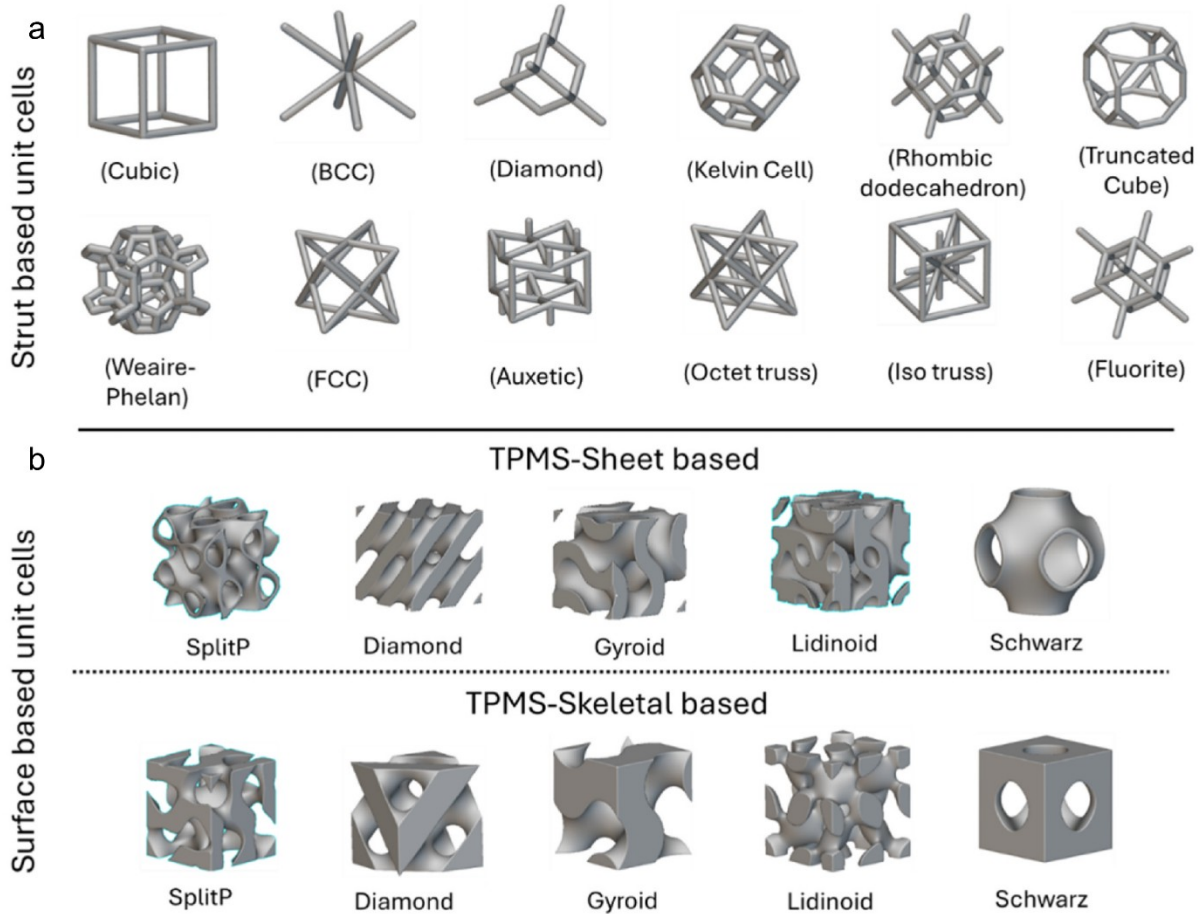


Figure 2-14- Various a) strut- and b) surface-based unit cells, defined for study of cellular lattice structures [149].

2.6.4. Cast aluminum lattice structure

As already mentioned, the production of metal lattice structures has been revolutionized by additive manufacturing. It enables complicated and lightweight designs that were previously impossible to achieve using conventional methods. However, direct metal AM processes such as LPBF face challenges in the production of large components, particularly in terms of cost, build time, printing defects, residual stresses, and material limitations. As an alternative, the investment casting process can be used to produce high-quality metal lattice structures with greater material flexibility and lower production costs.

By using AM to produce sacrificial polymer patterns, complex lattice designs can be transferred smoothly into metal via investment casting. This hybrid approach

combines the design freedom of AM with the scalability and metallurgical advantages of casting, making it a promising way to produce lattice structures.

Although the process chain of hybrid investment casting is cost-effective and rather simple, certain optimizations are necessary to ensure high-quality production. The choice of printing material and AM process parameters plays a crucial role. For example, when using fused deposition modeling (FDM) to fabricate polymer patterns, key factors such as filament composition for burnout compatibility, nozzle diameter and layer height for fine resolution, and optimized support structures for complex geometries must be carefully considered [35]. Furthermore, dimensional inaccuracies can arise at both the macro- and meso-levels due to the layer-by-layer nature of AM, requiring compensation strategies to improve overall precision [36]. Some AM processes offer greater precision, adaptability, and efficiency for pattern production, making them more suitable for investment casting applications. Selecting the optimal AM technique can significantly enhance the accuracy, surface quality, and reproducibility of the final cast structures.

Beyond AM-related challenges, investment casting itself has design constraints that must be addressed. The gating and feeding systems play a crucial role in ensuring complete mold filling and minimizing casting defects [38]. The topology of the lattice structure significantly influences both printability and castability. Since design is a multi-objective optimization, it is essential to select the self-supported and open-celled lattice structures, while focusing on enhancing the cast process to improve the mechanical performance [38].

Despite its advantages, hybrid investment casting introduces specific defects that differ from those observed in directly printed metal lattice structures. Compared to LPBF-produced lattices, which often exhibit granular surface textures due to partially fused metal powder, cast lattices are more susceptible to shrinkage porosity, particularly in thicker sections [42]. Additionally, differences in solidification rates result in significantly different microstructures. Directly printed metal lattices tend to have finer grains, which enhance mechanical properties, whereas cast lattices often develop coarser grains and intermetallic phases that can reduce strength and ductility [43]. The combination of casting defects and an unoptimized microstructure contributes to the brittle nature of cast lattice structures. Consequently, even when produced with the same geometry, lattices manufactured via casting and LPBF can exhibit fundamentally different deformation behaviors, impacting their suitability for various applications.

2.6.5. 3D-Printed aluminum lattice structure

While investment casting provides a feasible route for producing metal lattice structures, additive manufacturing (AM) has emerged as a more precise and flexible alternative, particularly for aluminum alloys. Direct metal AM processes enable the fabrication of complex lattice designs with high geometric accuracy and fine microstructural control [33]. Unlike cast lattices, which are prone to shrinkage porosity and coarse microstructures, printed aluminum lattices benefit from rapid solidification, resulting in refined grains and improved mechanical properties. Additionally, AM allows for the direct integration of optimized lattice topologies, enabling tailored performance for specific applications.

The production of aluminum lattice structures using direct additive manufacturing, specifically Laser Powder Bed Fusion (LPBF), presents unique challenges. While LPBF enables the fabrication of highly complex geometries that would be directly impossible with conventional methods, the interplay between process parameters and the complexity of the desired lattice geometry must be carefully managed. Unlike bulk components, lattice structures require the laser to frequently start and stop over short distances, leading to fluctuations in energy input that can result in inconsistent melting, balling effects, or partial fusion. Furthermore, inadequate process optimization can introduce defects such as lack-of-fusion porosity, keyhole porosity, or excessive surface roughness, which negatively impact mechanical performance [33,144]. The combination of high thermal gradients and localized heating can also lead to residual stresses, potentially causing warping or even microcracking, especially in aluminum alloys with high thermal conductivity [150].

Beyond the printing process itself, post-processing introduces further difficulties [31]. One significant challenge is the extraction of loose powder trapped within the internal cavities of the lattice, which is particularly problematic for designs with small strut diameters or closed-cell structures. Unremoved powder can alter the effective relative density and mechanical properties of the final part, potentially leading to unintended stress concentrations. Surface finishing is another concern, as LPBF-produced structures often exhibit high roughness due to partially fused powder particles adhering to the surface. This roughness can not only influence mechanical properties such as fatigue resistance but also hinder the functional integration of the lattice structure in load-bearing applications. Furthermore, heat treatment should be used to relieve residual stresses induced by rapid thermal cycling during printing, which can otherwise lead to distortion, reduced ductility, or premature failure under mechanical loading [151,152]. Addressing these challenges requires a combination of optimized scanning strategies, process parameter

refinement, and tailored post-processing techniques to enhance surface quality and structural integrity.

While the state of the art provides a foundation on the design principles, processing methods, and mechanical behavior of aluminum-based cellular materials, several gaps remain in understanding the possibility of microstructural improvement and the combined influence of microstructural and geometrical factors, especially in cast open-cell cellular structures. Most studies have focused either on optimizing the overall lattice geometry or on tailoring the bulk alloy properties, with limited attention given to how local microstructural features, such as dendrite morphology, phase distribution, or porosity, interact with the architecture to determine mechanical performance. Moreover, the majority of existing work emphasizes additively manufactured or powder-processed cellular materials, leaving cast structures, which may exhibit unique solidification-induced heterogeneities, less explored. This gap is particularly critical because the mechanical behavior of cast open-cell structures can be highly sensitive to fine microstructural variations, which may interact synergistically with other structural levels. A more comprehensive understanding of these interactions would not only enable better prediction of deformation mechanisms and failure modes but also provide guidelines for combined alloy design and structural optimization.

3. Research Hypothesis

Despite the widespread production of aluminum foams, due to their unique structural and functional advantages, a systematic approach to optimizing their properties based on mechanical performance has yet to be fully explored. While general understanding of aluminum alloys in bulk castings provides some insight, it cannot be directly applied to the open-cell foams produced by the block mold casting route due to the specific challenges arising from their solidification characteristics. Therefore, a detailed investigation into both coarse and fine variations in alloy composition and foam microstructure is essential to develop a deeper understanding and enable targeted optimization.

Although open-cell aluminum foams produced by investment casting represent only a small fraction of the total production of these advanced materials, studying their solidification and microstructure offers valuable insights into the complex interactions between alloy composition, solidification phenomena, and microstructure evolution under non-conventional casting conditions. These materials undergo unique solidification characterized by low cooling rates and restricted solidification spaces, which distinguish them from conventional bulk castings. Under these conditions, the development of microstructural features differs significantly from what is typically observed. As a result, the established principles of microstructural evolution in bulk castings cannot be directly applied to open-cell foams. This makes it critical to explore customized approaches to enhance their microstructure and, consequently, their mechanical properties. By adjusting alloy composition and refining casting parameters, it is aimed to achieve improvements in the performance of stochastic aluminum foams. This study not only provides insights into the fundamental solidification mechanisms under restricted geometries, but also attempts to close the knowledge gap to optimize the alloy composition and casting processes of open-cell aluminum foams.

To achieve this, the first step is to gain a deeper understanding of the block mold casting process in production of open-cell foams. In addition to the usual challenges encountered in casting processes, such as inadequate melt quality, shrinkage, inconsistent cooling, and various casting defects, block mold casting introduces further complexities. These include maintaining precise dimensional accuracy during mold preparation and achieving consistent dimensional stability in the cast counterpart. Overcoming these challenges is essential for improving the production of aluminum foams and ensuring that the desired cellular structures are consistently achieved.

With an understanding of the key parameters that influence the microstructure and properties of open-cell cast foams, the same principles can be extended to lattice structures produced under the same casting process parameters. However, in contrast to open-cell stochastic foams, there is an additional critical factor in lattice structures: the geometry and design of the unit cells. Realizing the potential of design innovation in lattice structures requires advanced manufacturing technologies to meet the complex requirements. Modern advancements in additive manufacturing techniques have revolutionized the possibilities for production of these complex structures, enabling precision and adaptability. Establishing a process chain to use sacrificial patterns from additive manufacturing in investment casting can offer an innovative process for the production of aluminum lattice structures. The integration of additive manufacturing with investment casting for cellular material production combines the design freedom of AM with a deep understanding of the casting process, unlocking new possibilities for material and properties optimization.

The arrangement, size, and shape of the lattice cells significantly impact the mechanical behavior, stress distribution within the struts, and the overall performance of the lattice structure. These geometric considerations interact with the underlying microstructural features, creating a complex relationship between design, alloy concept, microstructure, and properties. By applying the knowledge gained from the study of stochastic foams, it is possible to optimize lattice structures performance. This approach highlights the flexibility of the casting process and its potential for the development of cellular materials with precisely controlled properties. In addition, a comparison can be made between the manufacturing processes of lattice structures, namely direct additive manufacturing and hybrid investment casting.

With advances in additive manufacturing, this process is expected to dominate in future for the production of complex parts such as lattice structures. In the scope of this research, it can be realized that the alloy development based on the designed geometry can initially be carried out using the more established and convenient casting process. After initial characterization, the developed alloy concept can be employed in additive manufacturing.

4. Experimental Procedure

4.1. Sample production

4.1.1. Pattern

Two main sample systems were used for the purpose of this dissertation. For the experiments with stochastic foams, 10 ppi stochastic polyurethane (PUR) foam blocks were supplied from FOAMPARTNER Reisgies Kunststoffe GmbH (Leverkusen, Germany) with a height of 50 mm. According to the manufacturer, the average cell size of the foam is 4500 μm at 10 ppi. The PUR foam was cut to the desired dimensions using a hot wire foam cutter based on the test plan and the required analysis (Figure 4-1a). The foam samples were glued to a 3 mm thick investment casting wax plate and then connected to the rest of the casting system (pouring funnel and runners) out of wax as Figure 4-1b and c. All wax parts are manufactured using a manual wax injection device, wax A7-11 (Blayson Casting Systems Ltd, Cambridge, United Kingdom), and corresponding aluminum matrices.

The second set of patterns is considered for the experiments on lattice structures. In this study, the software nTop (formerly nTopology, New York, USA) was used to design the lattice structures. This is an advanced engineering software that enables the creation of lattice structures, topology optimization and additive manufacturing support. It uses implicit modeling and field-driven design to create complex geometries with high precision. The software provides tools to create and customize different types of lattice structures to match mechanical properties to specific requirements. nTop also optimizes material distribution within a design space to achieve specific performance goals while minimizing weight. By using nTopology, the lattice structures were efficiently designed within a cube shape with a length of 45 mm. After the complete design of each structure in ntop, a mesh was created from the designed implicit body with the tolerance of 0.3 mm. The meshes were used to export the STL files to be used for 3D printing of the patterns.

The 3D printer Perfactory 4 mini XL (EnvisionTec GmbH, Gladbeck, Germany), which operates using digital light processing (DLP) technology, was employed for this work (Figure 4-1d). To ensure high accuracy in producing small lattice structures and to achieve a smooth surface finish, a 75 mm lens was used, limiting the build area to $84 \times 52.5 \text{ mm}^2$. The projector brightness was set to 750 mW/dm^2 . Before each print, the compensation mask was checked across 36 points within the build area and adjusted if measurements deviated from the $\pm 5\%$ tolerance. This compensation mask ensures uniform light intensity across the entire build platform, which is essential for consistent part quality.

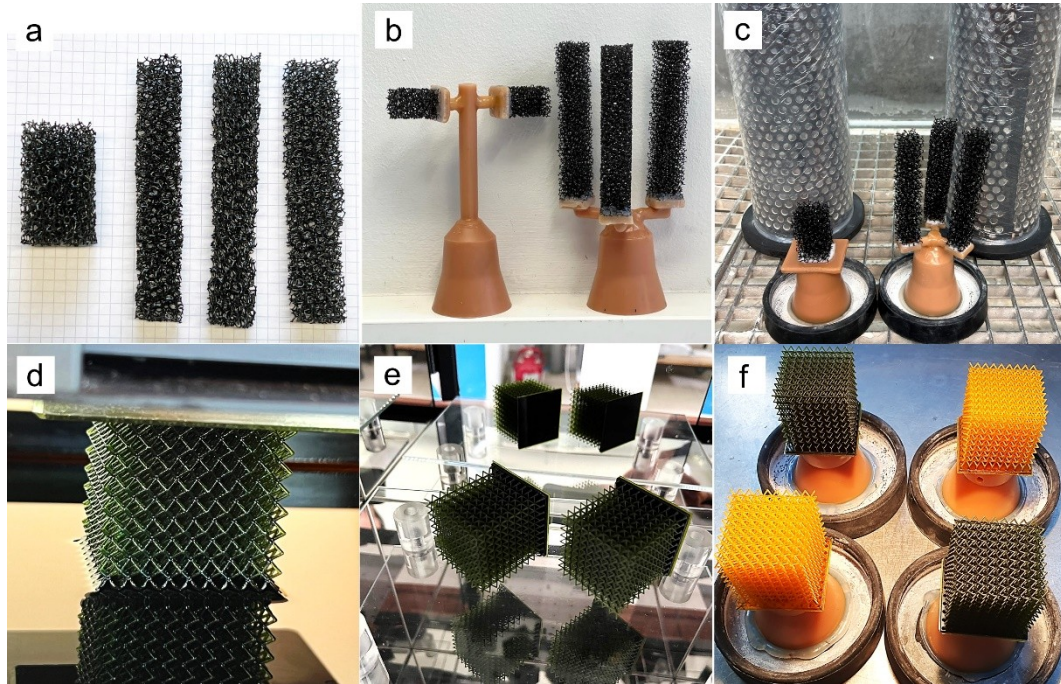


Figure 4-1- Pattern system for production of stochastic foams: a) cutting of the PUR foam to the desired dimensions, b) connecting to the wax casting system, c) preparing for block molding. Pattern system for production of lattice structures: d) 3D-printing of patterns using VAT polymerization method, e) final curing in a UV box, f) preparing for block molding.

Patterns used in investment casting must burn out cleanly, leaving minimal residue (ash). Standard resins used in vat polymerization often contain fine solid particles to enhance strength, but making them unsuitable for casting applications due to the excessive residue. To address this, manufacturers of 3D printers and materials offer specialized casting resins with a reduced filler content. While these casting resins result in patterns with lower mechanical strength, they ensure a clean burnout with less than 1% residue.

According to the resin products from EnvisionTec, Epic green and PIC 100 (orange color) are suitable resins for fine jewelry casting and support printing build styles with voxel thicknesses of 25, 35, and 50 microns. These resins are formulated from thermoset monomers, synthetic wax, and photo-initiators. Preliminary testing demonstrated a low residual ash content of 0.05 wt% for Epic Green and 0.06 wt% for PIC 100, along with excellent reproduction of fine structures, stability during mold preparation, and low thermal expansion during burnout. These characteristics help minimize the risk of mold cracking during the firing process.

After each print, the pattern was removed from the build plate and thoroughly washed with isopropanol to eliminate any residual resin. It was then stored at room temperature for 24 hours to ensure complete drying. Final curing was performed in

4. Experimental Procedure

an ultraviolet (UV) chamber equipped with internal mirrors, where each side of the cube was irradiated for 2 minutes (Figure 4-1e). Following curing, the printed patterns were affixed to the wax funnels (Figure 4-1f).

4.1.2. Molding

Block mold casting process as the modified investment casting route [130], was employed for the production of aluminum cellular structure in this research [44,45,153,154]. Patterns were mounted in covered, perforated steel flasks to be embedded in the investment molding material (Figure 4-2a). A gypsum-bonded investment slurry was prepared by thoroughly mixing Goldstar XXX investment powder (Goodwin Refractory Services Ltd., Staffordshire, UK) with water in a weight ratio of 100:40. To remove the air entrapped during mixing, vacuum degassing was carried out in two stages: first after mixing, and then again after pouring the slurry into the flasks. During the filling process, gentle vibration was applied to enhance slurry penetration into the cellular structures and to facilitate air release (Figure 4-2b). The timing of each step in the molding process was precisely controlled in accordance with the manufacturer's specifications for the investment powder. In tests requiring uniform cooling in the longitudinal direction, the slurry was poured into the cuvettes to a height of at least 50 mm above the patterns.

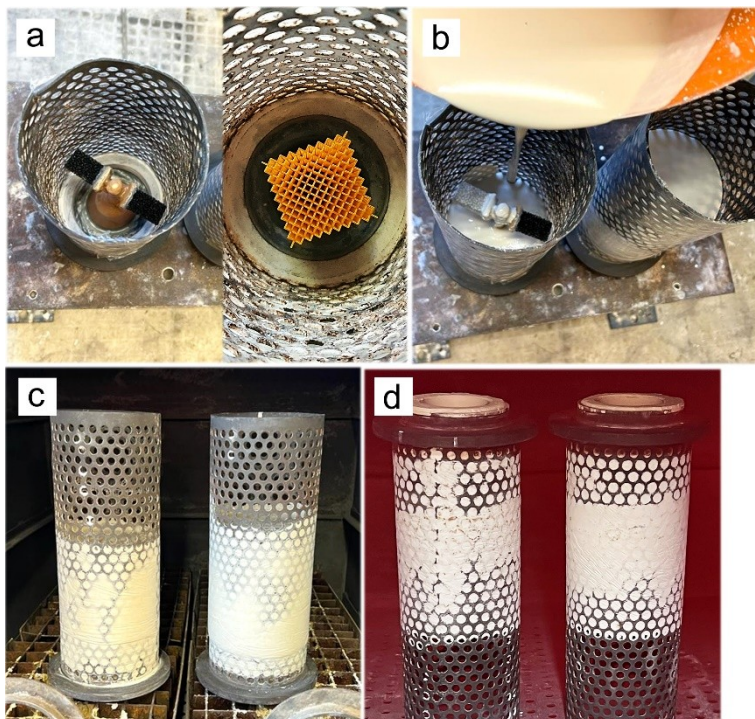


Figure 4-2- Block mold casting as the modified investment casting route: a) Patterns mounted in covered, perforated steel flasks, b) gypsum bonded investment slurry poured over the patterns, c) final drying and dewaxing inside the oven, d) burning of patterns and molds.

After a primary drying period of 2 hours at room temperature, final drying and dewaxing were carried out at 135 °C for 4 hours in an air-circulating oven (Figure 4-2c). From the beginning of molding until this point, the flasks were placed upside

down; however, after dewaxing, they were positioned straight up (the pouring basins faced upwards) to be ready for casting (Figure 4-2d). Burning of the PUR foam and firing of the ceramic molds were done at 720 °C in the same air-circulating oven. In the case of printed patterns, another step at 400 °C was added to the heating cycle and the final burning step was done at 750 °C. Figure 4-3 shows the entire heat cycle applied on the molds after initial setting of the investment slurry.

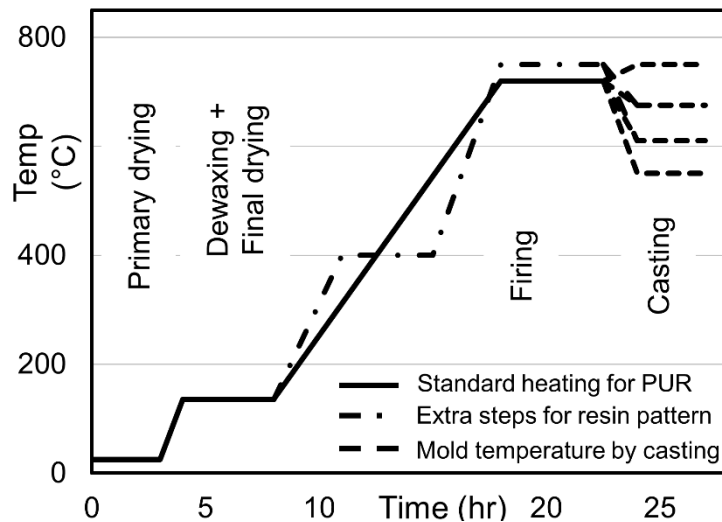


Figure 4-3- Heating cycle applied to the molds after filling with the molding material.

4.1.3. Casting

Depending on the targeted chemical composition to be cast, the input materials for the melt were determined and the corresponding quantities calculated. For AlSi7Mg, the pre-refined, unmodified alloy A356.0 from Aluminum Rheinfelden Alloys GmbH (Rheinfelden, Germany) was used without any further additions. This alloy (A356.0) also served as the base material for the production of AlSi10Mg, with the required amount of AlSi50 master alloy added to achieve the desired silicon content. For the alloys containing strontium and europium, appropriate quantities of AlSr10 master alloy (HOESCH Metallurgie GmbH, Niederzier, Germany) and AlEu5 were added to the melt at 700 °C. For the Al-Cu and AlSi3Mg alloys, commercially pure aluminum ingots (99.8% Al) were used as the base material, to which the necessary master alloys of AlSi50, AlCu20, and AlMg10, were added. In addition, the required amount of manganese was introduced using the AlMn20 master alloy.

All input materials were melted in a medium-frequency induction furnace, with a maximum melt weight of 3 kg to ensure rapid melting, thereby minimizing gas absorption and oxidation. To promote effective absorption of alloying elements and modifiers, the melts were heated to 800 °C and held for 12 minutes. Consequently, multiple melts were prepared for each alloy composition to produce all required specimens. The resulting chemical compositions were analyzed using a Hitachi

4. Experimental Procedure

OE750 optical emission spectrometer, as shown in Table 4-1. For each alloy, the reported composition represents the average of multiple melts and multiple measurement points together with standard deviations.

Table 4-1- Average chemical compositions of the alloys used in this study, measured by optical emission spectroscopy. Values in parentheses indicate the standard deviation based on multiple melts and measurement points

Alloys	Si wt%	Fe wt%	Cu wt%	Mg wt%	Ti wt%	Mn wt%	Sr (ppm)
Commercially Pure Al	0.038 (0.002)	0.067 (0.003)	0.001	0.002	0.003	0.001	---
AlSi3MgMn	3.04 (0.015)	0.079 (0.006)	max. 0.002	0.31 (0.027)	0.11 (0.005)	0.29 (0.002)	---
AlSi7MgMn						0.30 (0.051)	---
AlSi7Mg							---
AlSi7MgSr200	7.08 (0.071)	0.092 (0.003)	max. 0.002	0.29 (0.013)	0.10 (0.034)	max 0.002	214 (53)
AlSi7MgSr400							443 (28)
AlSi7MgSr800							828 (33)
AlSi10MgMn						0.29 (0.068)	---
AlSi10Mg							---
AlSi10MgSr200	9.80 (0.115)	0.088 (0.012)	max. 0.002	0.27 (0.038)	0.11 (0.015)	max 0.002	214 (53)
AlSi10MgSr400							443 (28)
AlSi10MgSr800							828 (33)
Al-4Cu	0.049 (0.012)	0.079 (0.002)	3.94 (0.06)	---	---		---
Al-5Cu	0.041 (0.005)	0.083 (0.005)	4.91 (0.06)	---	---		---

Achieving proper mold filling is highly dependent on the casting conditions. When producing complex foam and lattice structures with ultra-thin struts, specific precautions are required to prevent premature solidification of the melt before the mold is completely filled. In addition to adjusting the mold temperature, that will be discussed later, the perforated flasks were subjected to a vacuum of -0.6 bar during casting using a vacuum chamber. This negative pressure helped draw the melt into

the fine channels of the mold cavity (Figure 4-4a). After casting, the molds were cooled to room temperature in ambient air and then cleaned using water at a pressure of 4 bar (Figure 4-4b). Final removal of the gypsum-bonded molding material from inside the structures (Figure 4-4c) was performed using a plaster remover solution (Ernst Hinrichs Dental GmbH, Goslar, Germany).



Figure 4-4- a) Casting with the help of under pressure, b) casting products after demolding using water stream, and c) final cast specimens after complete removal of the molding material.

4.1.4. Direct additive manufacturing of lattice structures

The laser powder bed fusion (LPBF) process was employed to produce lattice structures with a body-centered cubic (BCC) design using the AlSi10MgMn alloy in two separate batches. This alloy is commonly used in LPBF due to its good printability and mechanical properties, making it ideal for comparing cast and printed counterparts. The nominal chemical composition of the used powder is presented in Table 4-2. The powder size range was 20 to 63 microns, with a maximum of 10 vol% of particles smaller than 20 μm and a maximum of 10 vol% larger than 63 μm (Carpenter Technology Corporation, Pennsylvania, USA).

Table 4-2- Nominal chemical composition (in wt%) of the used powder for the LPBF production route

Al	Si	Fe	Mn	Mg	Ti	O
Balance	9.00-11.00	0.55	0.45	0.2-0.45	0.15	0.10
Zn	Cu	Pb	Ni	N	Sn	
0.10	0.05	0.05	0.05	0.05	0.05	

Various parameters such as laser settings, scanning strategy, build orientation, and post-processing conditions must be optimized to achieve the best performance. As a general rule, the layer thickness is typically set to 50–75% of the D90 particle size and based on previous experience, it was set to 40 microns for this study. A lower

layer thickness provides better resolution but increases printing time, while thicker layers may result in incomplete melting and poor surface quality. The hatch spacing was set to 100 microns.

In the first batch of printing, laser power of 200 W and scanning speed of 1200 mm/s were chosen for printing of BCC structures and in the second batch, laser power of 250 W and scanning speed of 1100 mm/s were chosen. To prevent distortion, the cooling rate during both heating and cooling cycles was controlled at approximately 10 °C/min. Furthermore, to reduce residual stresses in the printed lattice structures, a stress-relief heat treatment was applied to the second batch at 350 °C for 3 hours.

4.2. Analysis

4.2.1. Thermal analysis and temperature measurement

Type K thermocouples (NiCr/NiAl) with a wire diameter of 0.2 mm and a total tip diameter of 0.5 mm (TC Direct, Mönchengladbach, Germany) were used in conjunction with a temperature recording system to perform thermal analysis on the melt. Temperature measurements were taken in the pouring basins to determine the liquidus and eutectic temperatures of the used alloys. The pouring basin, which is funnel-shaped due to the wax gating system, provided a sufficient amount of melt for thermal analysis.

The same temperature measurement system was also used to monitor the temperature and cooling rate at specific locations within the foam structure. A thermocouple tip was fixed inside the cells at predefined positions using superglue, ensuring it remained secured during molding. While the thermocouples measured the temperature of the investment material inside the foam cells, direct measurements from the metal in the thin struts were not feasible. However, since the struts are fully covered by the hot molding compound, these temperature measurements effectively represent the cooling behavior of the foam at the thermocouple positions. To estimate the cooling rate, linear cooling between the liquidus and solidus temperatures was considered.

4.2.2. Microstructure study

The strut microstructure, including the existence and distribution of aluminum dendrites, eutectic silicon phase, intermetallic phases, and precipitates, was investigated on unetched metallographic samples. These samples were embedded in the conductive cold embedding material KEM 70 (ATM Qness GmbH, Mammelzen, Germany). The study surface was prepared for microscopic investigation by grinding with SiC abrasive papers up to a 1200 grit finish, followed by polishing with diamond solutions of 9, 3, and 0.025 µm particle sizes using a grinding and polishing

machine (Saphir 550, ATM Qness GmbH) as outlined in Table 4-3. As the final preparation step, the samples were polished with a colloidal silica suspension ($0.02\ \mu\text{m}$) on a VibroMet polishing machine (Buehler-ITW Test & Measurement GmbH, Düsseldorf, Germany). Optical micrographs were captured using an Axio Imager A1 microscope (Carl Zeiss, Oberkochen, Germany).

Table 4-3- Metallographic samples preparation steps

Rotation direction	Rotation speed	Action	Load (N)	Characteristic	Duration (min)
Counter rotation: Sample holder clockwise	300 rpm: Sample holder 150 rpm	Grinding by SiC abrasive paper	15	180 grit	2
				320 grit	2
				500 grit	3
				1200 grit	3
Polish disk counterclockwise	Polish disk 150 rpm	Polishing by diamond solution	20	$9\ \mu\text{m}$ particles	5
				$3\ \mu\text{m}$ particles	5
				$0.025\ \mu\text{m}$ particles	3

During metallographic preparation, the struts from the stochastic foams were sliced in different directions. By changing the direction from cross-section to longitudinal section, a larger surface area could be examined. The microstructure examination was performed on at least 12 struts for each condition. These struts were chosen from different directions and locations of the prepared sections to provide a comprehensive understanding of the microstructural properties. The size, shape, and position of silicon particles were analyzed for at least 2000 particles per condition using the Fiji implementation of the open-source software ImageJ. For the lattice structures, one sample from each condition was analyzed in the same manner as the foam samples to ensure consistency in microstructure examination.

To make the aluminum grains with different crystallographic orientations recognizable, metallography samples were anodized using Barker's reagent, composed of 5 ml of fluoroboric acid (HBF_4) in 200 ml of water [155]. The anodizing process was carried out for 80 seconds using a direct current with an electric potential difference of 20 V. The anodized samples were then examined under polarized light using the same Axio Imager A1 microscope.

To observe the three-dimensional structure and morphology of the eutectic silicon phase, as well as the relative arrangement of aluminum and silicon on the strut

4. Experimental Procedure

surface, foam samples without embedding were deep-etched in a solution consisting of 15 ml hydrochloric acid (HCl) and 10 ml hydrofluoric acid (HF) in 90 ml of water for 2 minutes. Deep etching dissolves the aluminum matrix from the surface layer, leaving the eutectic silicon and intermetallic phases undamaged on the surface. The deep-etched struts were then analyzed using scanning electron microscopy (SEM).

4.2.3. Scanning electron microscopy and energy dispersive X-ray spectroscopy

Analysis of the fractured struts after the mechanical properties tests and on the etched surface of the struts was conducted using a scanning electron microscope (SEM) from Carl Zeiss (Oberkochen, Germany). The SEM was operated in secondary-electron mode, backscatter mode, or a combination of both, with the accelerating voltage typically set to 15 kV. The analysis of the chemical composition of the particles and precipitates, the line scan analysis and the element mapping were carried out with energy dispersive X-ray spectroscopy (EDX) within the same SEM device.

4.2.4. Compression test of cellular samples

A quasi-static uniaxial compression test was performed on foam specimens using a tension/compression loading stage from Kammrath & Weiß GmbH (Dortmund, Germany), as shown in Figure 4-5. The testing stage was equipped with a 500 N load cell that had a resolution of 0.02%, as well as a linear displacement gauge with a range of 12 mm and a resolution of 0.01%. During the test, the force was applied along the elongated cell axis at a rate of 5 $\mu\text{m/s}$.



Figure 4-5- A sprinkled foam specimen fixed between the compression plates.

To compensate for minor inaccuracies in the sample dimensions and to ensure that both sides of the foams were in full contact with the compression plates, the foam samples were preloaded with a small force between 0.2 and 0.5 N before starting the force and displacement recordings. To evaluate the strain fields with Digital Image Correlation (DIC), the surface of the specimens was sprinkled with very small

dots of black paint. Images were taken every 20 seconds during the compression test using a Sony DSC-RX10M2 digital camera (Sony Corporation, Tokyo, Japan). The image sequence was then analyzed using VIC2D software (Correlated Solutions Inc., Irmo, USA).

Five foam samples of each condition were tested under compression. From the nominal stress-strain curves, the following mechanical properties of foams were determined according to ISO 13314:2011 [77], and the average values are reported:

- i. Quasi-elastic gradient, m : The slope of the linear region at the beginning of the nominal stress-strain curve.
- ii. Peak stress, σ_{peak} : The maximum stress as the transition from the quasi-elastic to the plastic deformation state.
- iii. Plateau stress, σ_{pl} : The average stress between 20 % and 30 % strain.
- iv. Densification strain, ε_0 : The corresponding strain to the stress 1.3 times the plateau stress.
- v. Energy absorption per volume, W : Calculated from the area under the stress-strain curve until densification strain (Equation 4).
- vi. Energy absorption efficiency, Φ : The ratio of W to the product of the maximum stress σ_m and densification strain ε_0 (Equation 5).

To account for the density difference between the specimens, the stress values for each compression test were normalized by dividing the nominal stress (force divided by the specimen's cross-sectional area) by the relative density of each specimen. The relative density was determined as the ratio of the foam density to the bulk alloy density (e.g. for A356, $\rho_{solid} = 2.66 \frac{g}{cm^3}$ [18]). The foam density was calculated as the ratio of the sample mass to the overall sample volume.

For the lattice structure, the compression test was conducted on three samples for each condition using a general tensile/compression testing machine from Hegewald & Peschke Meß- und Prüftechnik GmbH (Nossen, Germany) with a maximum force of 100 kN, and average values are reported. The compression speed was set to 3 mm/min. The test began when the force exceeded 5 N and ended when 80% of the compression was reached, corresponding to a height reduction of 35 mm. A camera was positioned at the center of the sample to capture photos of the compression process every 20 seconds.

For the mechanical properties of the lattice structure, the plateau stress was calculated as the average stress between the peak point and 50% elongation. The densification strain was defined by the intersection of the tangents to the stress

4. Experimental Procedure

plateau regime and the densification regime [75] and the remaining mechanical properties were determined in the same way as for the foam samples.

4.2.5. Micro-tensile test of single struts

Individual struts were carefully extracted from the foam samples of different conditions, selected so that they were straight and free of visible damage and were subjected to micro-scale tensile testing [156,157]. Five to seven struts per condition were tested under tensile load using the same Kammrath & Weiss micromechanical loading stage employed for the foam compression tests. The displacement speed during the testing was set to 1 $\mu\text{m/s}$.

To conduct the tensile test on the tiny struts without causing any damage from the custom-made grips, both ends of the struts were embedded into Filtek Supreme dental restorative material (3M Deutschland GmbH, Germany). The resin was applied to both grips as shown in Figure 4-6. The strut was carefully placed into the soft/liquid resin and aligned under the microscope. Finally, the resin was hardened using UV light, securing the strut position for testing.

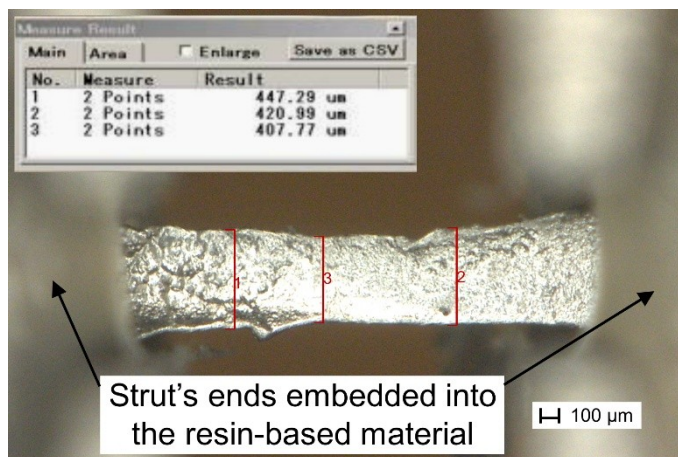


Figure 4-6- Embedded single strut into the dental resin-based material for tensile test [45].

For strain determination, micrographs of the specimen surface were captured every 15 seconds using a VHX-100K microscope (Keyence Deutschland GmbH, Neu-Isenburg, Germany). The photo sequences and the movement of reference points, as close as possible to the grips, were analyzed through the virtual extensometer function of the VIC2D software. This allowed for precise tracking of strain during the tensile testing of single struts.

To calculate the stress for this analysis, both the force and the cross-sectional area of the individual struts are required for each test. The force is measured using the loading device, but measuring the cross-sectional area of the struts is more complex due to their concave triangular shape and varying cross-sectional area along the longitudinal axis. Since it is difficult to measure these complex cross-sectional areas for each individual strut, an estimation strategy was employed [72,156].

More than thirty struts were extracted from the fully filled foam specimen and mounted upright in cold embedding material. The strut cross-sections were prepared approximately in the middle of the strut length as in a standard metallographic preparation. For each strut, a triangle was approximated, matching the corners of the strut surface, as shown in Figure 4-7. The edge lengths of the triangle (a, b, and c) and the height of the circular section (h), which represents the difference between the area of the triangle and the area of the strut, were measured. Based on these values, both 2D and 3D regression analyses were performed.

The 2D analysis provided an estimated equation linking h to the mean value of the edge length (x), while the 3D analysis provided an equation relating the radius of curvature, h, and the mean edge length. The high coefficients of determination (R^2 values) for these regressions (86% for the 2D regression and 80% for the 3D regression) indicate that the equations offer a good fit to the data, making them reliable for estimating the strut cross-sectional area.

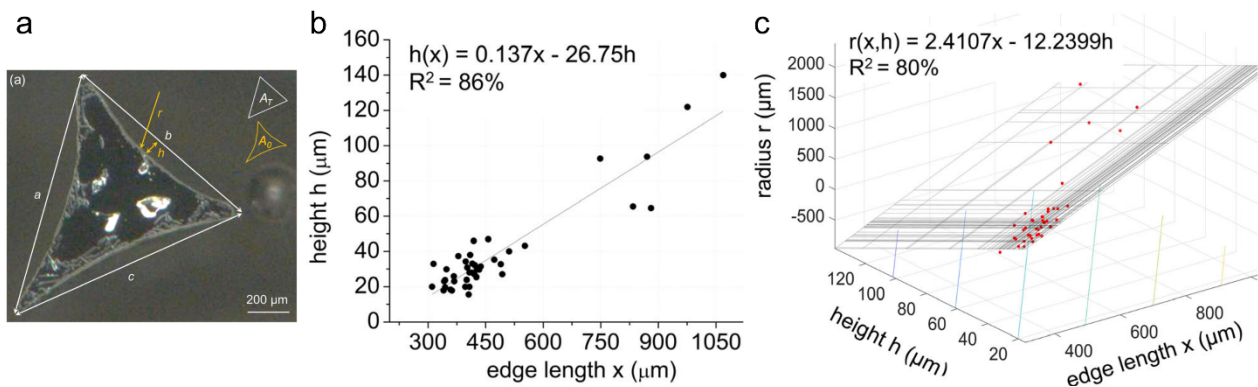


Figure 4-7- a) Light micrograph of a typical strut cross section, approximated by a triangle (white arrows) corresponding to the projected width of the strut. The “height” h as the distance between the circular section and the edge of the triangle, and the parameter r as the radius of the circular section, b) 2D regression between the edge length and the height, c) 3D regression between the curvature radius, the height, and the edge length [153].

Before each tensile test, all three edge lengths of each strut were measured using the Keyence VHX-100 digital light microscope (Figure 4-6). From the average value, the estimated values for h, r and also the cross-sectional area for each strut were determined. The cross-section area was then used to calculate the stress from the force measurement.

4.2.6. Micro-hardness measurement

The micro-hardness of Al-Cu foams was measured in both the as-cast and aged conditions using Vickers indentation with an ATM Qness device (model 60 M Evo-ATM Qness GmbH) and a force of 0.025 Kgf. For this measurement, struts were

extracted, embedded vertically, and polished. Indentations were performed on polished cross-sections of approximately six struts per condition.

4.3. Microsegregation model for Al-Cu alloy

The strengthening mechanisms of Al-Cu alloys, namely precipitation and solid solution hardening, can be impaired by microsegregation during solidification. Furthermore, microsegregation affects the homogenization and age-hardening responses in Al-Cu alloys, making it important to study. ThermoCalc 2022a, with the DICTRA module, was used to simulate microsegregation between the dendrite arms. The TCAL5 thermodynamic and MOBAL4 mobility databases were employed to perform the simulation. The secondary dendrite arm spacing (SDAS) and the cooling rate, which are crucial inputs for the simulation setup, were determined from microscopic images and the cooling curve, respectively.

Investigation of Cu microsegregation during solidification and further cooling to room temperature was performed using a moving boundary multiphase system model. As Figure 4-8 illustrates, the simulation setup defined the computational domain as half of the secondary dendrite arm spacing (SDAS) in the mushy zone due to symmetry. The simulation began at a temperature of 700 °C with only the liquid phase present. As the temperature dropped to room temperature, the thermodynamic driving forces for phase formation were calculated using DICTRA. The α -Al and θ -Al₂Cu phases started to form at the left and right interface boundaries, respectively, when they became thermodynamically stable. A planar geometrical model was used to approximate the shape of the dendrites. Only Cu was considered during the calculations, and all other impurity elements and associated phases were excluded from the computational domain for simplification.

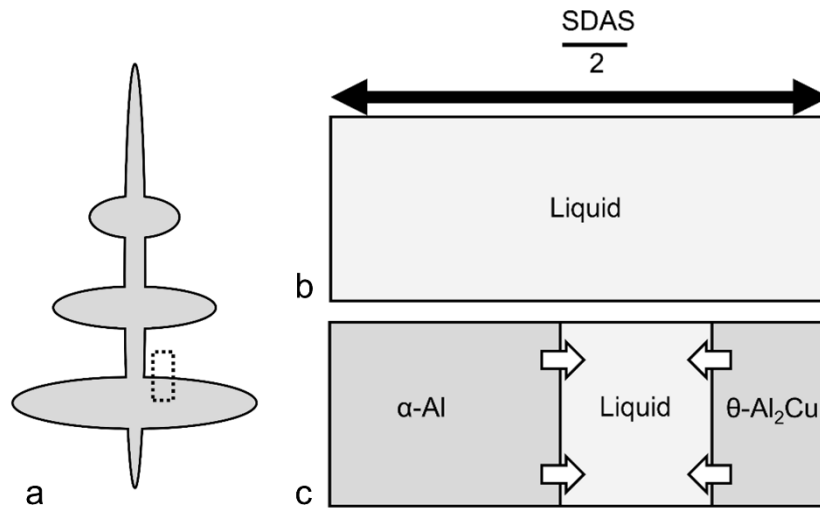


Figure 4-8- Schematic illustration of the simulation setup: a) the microsegregation domain (the dotted rectangle) defined as half of the SDAS, b) liquid phase stability above the liquidus temperature at the start of simulation, c) appearance of $\alpha-Al$ (at early solidification stage) and $\theta-Al_2Cu$ (at later solidification stage) phases at the domain boundaries.

5. Design of Experiments

To answer the research questions arising from the state of the art, the first phase of this dissertation attempts to better understand the block mold casting process in the production of open-cell aluminum foams. The casting parameters, influencing the microstructure of primary aluminum and eutectic silicon phases, are discussed in order to identify their effects on the mechanical properties and deformation behavior. In the second phase, the effects of the microstructural features, i.e. the coarse and fine changes of the secondary phases, on the mechanical properties are systematically investigated in order to understand the extent to which the mechanical properties of stochastic foams can be influenced through microstructure adjustments. The third phase employs the knowledge gained in the first two phases and applies it to the production of aluminum lattice structures. In addition to examining the influential parameters discovered in the earlier phases, the impact of cell geometry will be further explored. To begin the third phase, an indirect additive manufacturing process is going to be implemented to produce consistent aluminum lattice structures within the same route of the block mold casting process. Figure 5-1 summarizes the phases and steps tackled in this work.

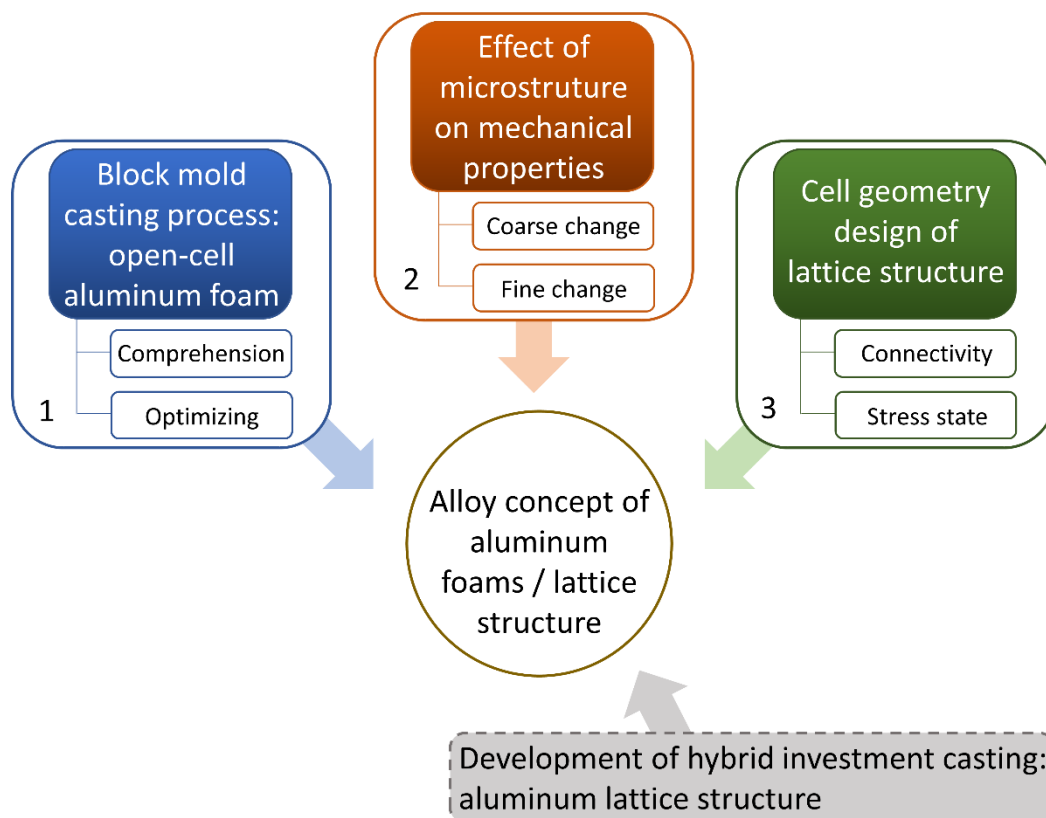


Figure 5-1- Overview of the different steps in this work to enable the study of the relationship between process, microstructure, and properties of cast aluminum foams and lattice structures.

5.1. Influence of the casting process on microstructure and properties

According to S. Fischer et al [79,130], mold and melt temperatures are the two most important parameters that influence the microstructure and mechanical properties of aluminum foams. For AlSi7Mg alloys, mold temperatures of 700 °C and 750 °C were found to be effective for successful castings. A. Matz et al. also employed the block mold casting process, using centrifugal force to fill the mold with lower mold temperatures of 400 °C and 600 °C [71,140]. Due to the slow cooling rates resulting from these casting conditions, the struts of investment cast foams are locally monocrystalline, with most struts containing a single α -Al dendrite surrounded by eutectics towards the strut surfaces.

To investigate the effects of the casting conditions on the microstructure and mechanical properties in more detail, 6 mold temperatures and 2 melt temperatures (Table 5-1) are used for production of AlSi7Mg foams. For each condition, two molds are cast with different sample dimensions. The first mold contains one sample type A for microstructural investigations, and the second mold contains three samples of type B (all connected to a main down sprue) for mechanical properties tests. All the information about the different sample types is summarized in Table 5-2 and Figure 5-2.

Table 5-1- Casting conditions defined by the mold and the melt temperatures for AlSi7Mg foams

Casting conditio	Mold temperature (°C)	Melt temperature (°C)	Casting conditio	Mold temperature (°C)	Melt temperature (°C)
1	750	700	7	600	700
2		740	8		740
3	675	700	9	590	700
4		740	10		740
5	610	700	11	575	700
6		740	12		740

During the preliminary testing phase and while casting long foam samples without changing the casting conditions, variations in the microstructure were observed along different sections of the samples. To investigate the influence of cooling conditions on the microstructure in the longitudinal direction, type C foam samples (Figure 5-2c) were produced in the center of the casting flasks. Temperature

5. Design of Experiments

measurements are taken at three different sections of type C sample at the center of the foam: longitudinal section 1 in the vicinity of the pouring basin, longitudinal section 2 in the middle of the sample length, and longitudinal section 3 at the end of the foam, to assess the consistency of cooling along the length of the sample.

Table 5-2- Dimensions, locations, and purposes of samples used in the different investigations of this work

Sample designation	Cast dimension (mm)	Specimen dimension (mm)	Location	Purpose
Type A	30×30×60	30×30×25	Middle of the flask	Microstructure investigation at different casting conditions
Type B	20×20×120	20×20×20	3 samples near the flask center	Mechanical properties tests at different casting conditions
Type C	30×30×120	---	Middle of the flask	Microstructure and cooling study to check longitudinal consistency
Type D	20×20×35	---	Radial direction from the flask center to the flask wall	Microstructure and cooling study to check horizontal consistency
Type E	50×50×120	50×50×50	Middle of the flask	Mechanical properties tests of the heat treatable alloy concept
Type B-Plus	20×20×120	20×20×20	3 samples near the flask wall	Mechanical properties tests at higher cooling rate

Type D samples with dimensions of 20×20×35 mm³ are produced in a horizontal orientation along the radial direction of the casting cuvette (Figure 5-2d). This radial sample setup allows for an investigation of cooling homogeneity across the flask's cross-section, specifically at radial zones 1, 3, and 5, which represent positions from the center of the flask to near the flask wall.

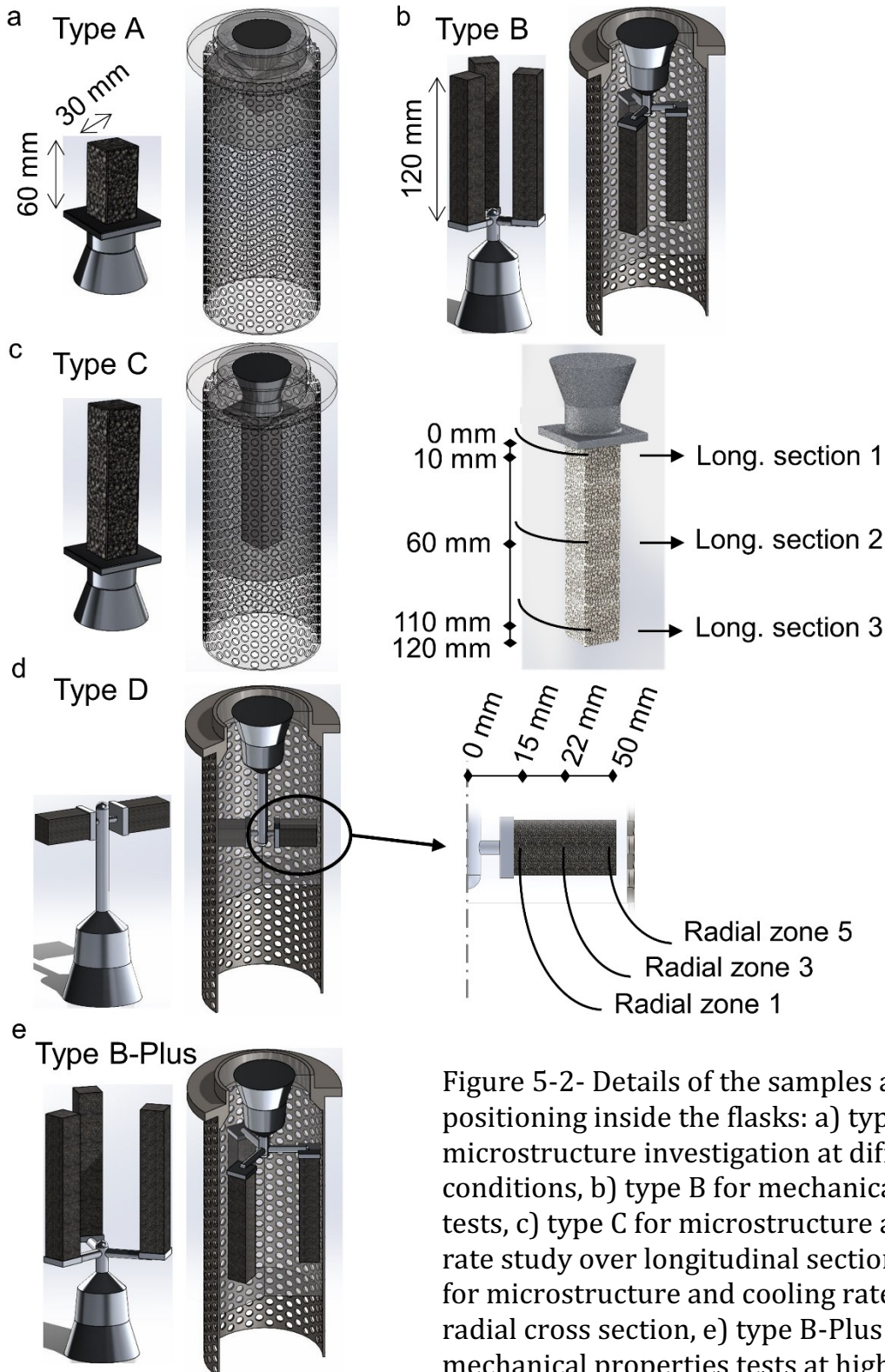


Figure 5-2- Details of the samples and positioning inside the flasks: a) type A for microstructure investigation at different casting conditions, b) type B for mechanical properties tests, c) type C for microstructure and cooling rate study over longitudinal section, d) type D for microstructure and cooling rate study over radial cross section, e) type B-Plus for mechanical properties tests at higher cooling rate.

The heat-treatable alloy concept (Al-Cu alloy) is tested using type E samples. This sample type is similar to type C but has a larger cross-section. In this case, the

samples for the mechanical property tests have dimensions of 50×50×50 mm, which enables a more comprehensive understanding of the properties.

To produce foam specimens for compression testing solidified at higher cooling rates, a modified version of the type B setup is introduced, as type B-Plus (Figure 5-2e). In this configuration, the three longitudinal samples are positioned close to the flask wall to expose them to enhanced cooling conditions. The adjusted positioning spans radial zones between positions 3 and 5.

5.2. Effect of secondary phase morphology

Eutectic silicon phase: Understanding the microstructure of the cast foams and their deviations from bulk counterparts will be helpful to investigate the influence of the different microstructural features on the mechanical properties. In Al-Si alloys, the microstructure typically comprises primary α -Al dendrites and Al-Si eutectic phases. However, in cast foams, the constrained geometry of the struts results in a distinct grain structure and altered eutectic distribution compared to bulk samples. The effects of the casting process on these two main microstructural components were investigated in the previous section, and the effects of the alloying on these characteristics is examined in this step.

Due to the low cooling rates during the solidification of open-cell foams, coarse, micro-sized silicon particles tend to form, which can significantly reduce the ductility of the alloy and influence the deformation behavior of the aluminum foams. Since a significant change in cooling condition is not feasible within this production route to modify the eutectic silicon phase, a chemical modification is employed to improve the size and distribution of the eutectic phase.

As known from the literature, the benefits of strontium addition for reducing silicon size and improving mechanical properties are diminished when the cooling rate during solidification is reduced. To address this, practical approaches involving slight adjustments to the cooling rate are tested in order to achieve a more homogeneous distribution of fine silicon phases across the strut cross-section. The gradient in cooling rate observed in the horizontal direction of the mold (sample type C) provides an opportunity to study microstructural changes at different cooling rates.

The effectiveness of alternative modifiers, such as sodium, strontium, and europium, is evaluated, and the most efficient one is selected for the final study on Al-Si foams. To determine this, the degree of modification is systematically evaluated for the chosen modifiers under various cooling conditions and at different concentrations. Cooling rates spanning from 0.05 to around 8 °C/s are achieved by casting bulk aluminum samples into molds with different thermal conductivities

and geometries. These include sand molds of varying thicknesses, gypsum-bonded molds with different preheating temperatures, one copper die, and steel dies with different wall thicknesses and preheating temperatures. Different casting conditions and corresponding cooling rates are summarized in Table 5-3.

Temperature measurements for each condition are carried out using type K thermocouples placed at the center of each sample, connected to a temperature measurement system. Characteristic eutectic solidification temperatures are determined from the cooling curves and their first derivatives, following the method of Tamminen [158]. These include the liquidus temperature T_{liq} (marking the onset of solidification), the eutectic nucleation temperature T_N , the minimum temperature before recalescence T_M , and the eutectic growth temperature T_G , as illustrated in Figure 5-3. Cooling rates were calculated as the slope between T_{liq} and T_G .

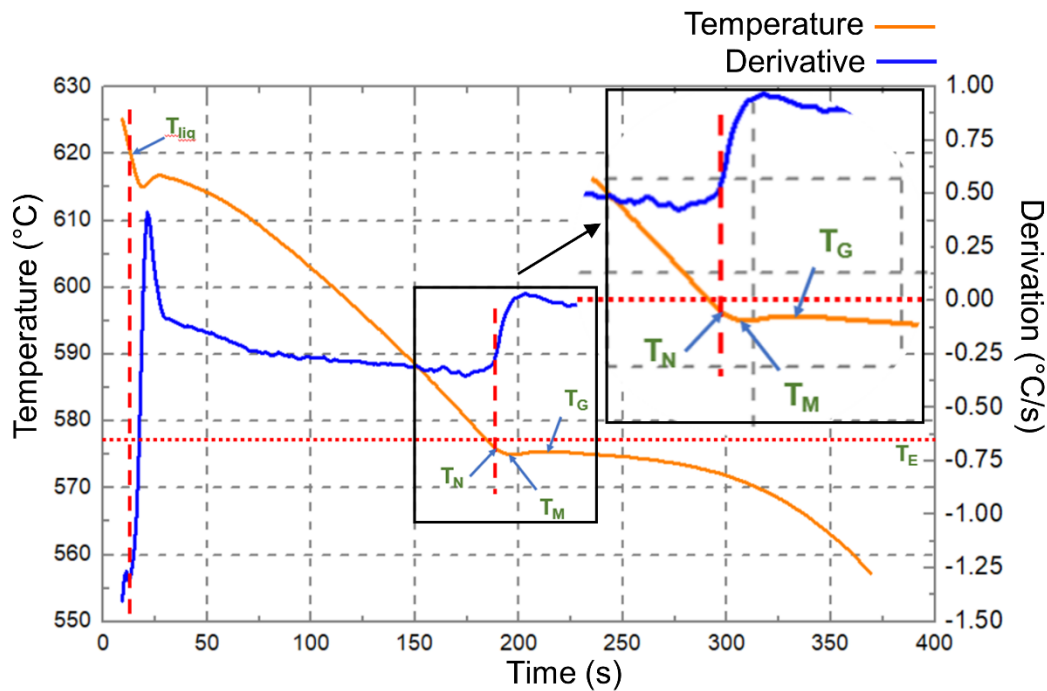
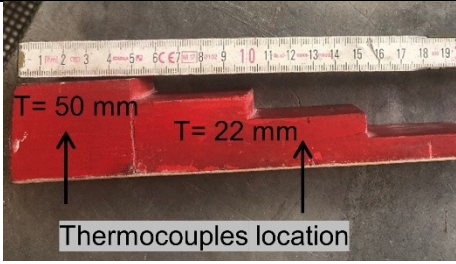
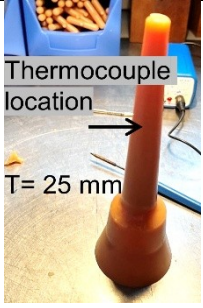
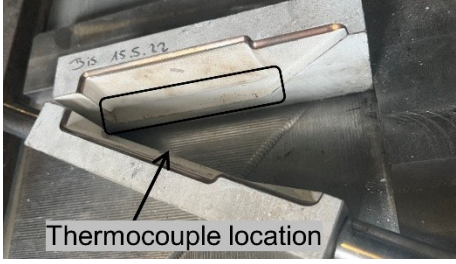
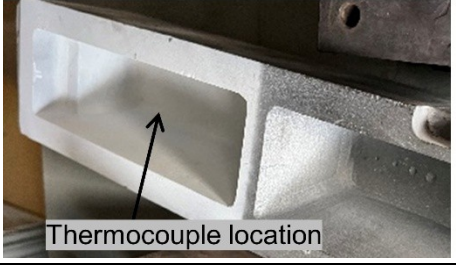
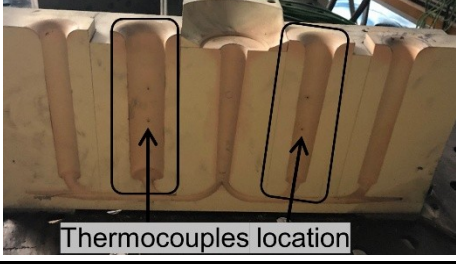


Figure 5-3- Example of a cooling curve and its first derivative, used to define characteristic solidification temperatures.

5. Design of Experiments

Table 5-3- Overview of mold materials, mold temperatures, and specimen thicknesses used to achieve a wide range of cooling rates for evaluating the modification efficiency of different modifiers

Mold material	Mold temperature (°C)	Thickness of specimen (mm)	Pattern or mold illustration	Cooling rate (°C/s)
Sand	25	50		0.24
		22		0.40
Gypsum	500	22		0.03
	300			0.11
	150			0.37
Copper	25	15		7.68
Steel	25	50		1.1
	500			0.2
Steel	25	25		3.59
		15		4.15

This study allows for a broad range of solidification conditions, including the cooling rates encountered in aluminum foam struts. After casting, the degree of modification is assessed as Equation 9 based on silicon morphology and eutectic refinement. Based on these findings, the optimal modifier is selected to ensure consistent microstructural refinement and improved mechanical performance in aluminum foams.

$$\text{Mod. degree} = \frac{\text{Ave. modified Si length} - \text{Ave. nonmodified Si length}}{\text{Ave. nonmodified Si length}} * 100$$

Equation 9

Iron intermetallic phases: Brittle iron rich intermetallic particles are typically found either in the primary aluminium α -phase or among the eutectic Si particles that influence mechanical properties and promote casting defects. Due to the limiting feeding possibility through the fine struts, accumulation of these particles can be a source for the shrinkage porosity in the foam castings. The low cooling rate during solidification can also negatively affect the morphology of the iron phases by promoting long acicular Beta phases. These particles can grow up to half the thickness of the strut size and greatly reduce ductility. The influencing factors of cooling rate and Mn addition are investigated to improve their morphology into Chinese script phases and a general solution to be suggested in order to reduce their negative influences.

Copper rich precipitates: The transition from coarse silicon particles to a fine, modified eutectic phase, as well as the transformation of β -iron phase morphology into compact, spherical particles, can effectively tailor the microstructure at the micrometer scale. To further investigate property enhancement, influencing the microstructure at the nanometer scale also presents an interesting opportunity. One approach that addresses both micro- and nanoscale structural changes is the addition of reinforcing particles of various sizes to the melt. This composite melt strategy can be advantageous when micro-sized particles with similar density and surface energy to the aluminum melt, such as Al_2O_3 or SiC, are used. These particles can be successfully mixed into the melt and entrapped within the structure during solidification.

However, the addition of nano-sized particles via an ex-situ route is generally impractical. Due to their high surface energy, nano-particles tend to agglomerate, leading to poor dispersion within the matrix. In addition, achieving strong bonding between these particles and the aluminum matrix is difficult, owing to mismatched surface properties. As a result, an in-situ approach is preferred to generate a range of reinforcing particles directly within the aluminum matrix. For this purpose, an

Al-Cu alloy is selected, as it can be readily heat treated to form precipitates of various sizes, providing a controllable and effective method to introduce nano-scale microstructural features.

Heat-treated Al-Cu components primarily gain their strength through precipitation hardening. After complete solidification, an annealing heat treatment is performed to dissolve any non-equilibrium phases that form during solidification. To achieve a homogeneous solution-annealed state, it is crucial to select the correct chemical composition to limit the presence of not-equilibrium copper-containing phases. This is especially important when the alloy solidifies at low cooling rates, as is the case in aluminum foam casting.

A detailed examination of the binary Al-Cu phase diagram reveals that the solubility limit of copper in aluminum at the eutectic temperature of 548 °C is around 5.65 wt.% Cu. While higher copper content (closer to this limit) can improve properties after aging, it also increases the risk of eutectic reactions during solidification, leading to brittleness due to the formation of coarse Al₂Cu (θ) phases. Even when the copper content is below the solubility limit, segregation during solidification can result in the non-equilibrium formation of θ -phase in the final regions to solidify. These as-cast precipitates can form in varying shapes and sizes and should be dissolved as much as possible before the aging treatment to maximize copper retention in the matrix.

At higher copper contents, the temperature window for solution annealing becomes very narrow, and if heating is not uniform, certain regions of the microstructure could re-melt. Therefore, it is generally better to stay below the 5.65 wt.% Cu limit. To overcome these issues, two binary alloys with 4 and 5 wt.% Cu are selected for this study. A simulation of copper atom segregation within the dendrite arm spacing in the as cast condition is conducted to evaluate the distribution of copper and understand the significance of solution annealing.

Foam samples of the size 50×50×50 mm³ are cast with the mentioned Al-Cu alloys from Table 4-1. Solution heat treatment is done for different times at 525 °C to study its influence on dissolution of Cu rich phases. Then, the foams are aged at temperatures ranging from 160 °C to 300 °C to generate precipitates of different sizes, from nanometer-scale Guinier–Preston (GP) zones to micrometer-sized θ -phase particles. The aging duration is fixed at 24 hours for all conditions.

5.3. Study of cell geometry design in lattice structures

Lattice structures with different cell geometries and the constant outer geometry of a cube with a length of 45 mm are produced by using the patterns from the vat

polymerization process within the same block molding casting method used for foam production. The first lattice structure is generated with the unit cell design of body centered cubic (BCC) with the unit cell size of 5 mm and strut diameter of 0.73 mm. A square base plate with the thickness of 3 mm and the length of 48 mm is considered at the bottom and cone shape supports are designed connecting the base plate to the lattice structure as Figure 5-4a. Considering the Al-Si alloy as the material, the software calculated the theoretical weight of the lattice structure as 25.02 g, which corresponds to a relative density of 10.17%.

The same design principle (unit cell size, base plate, and supports) was used to generate the lattice structures with Diamond and Kelvin unit cells keeping the relative density close to the one for BCC. Therefore, Diamond and Kelvin lattice structures with the respective strut diameter of 0.74 mm and 0.68 mm are generated (Figure 5-4b and c). The theoretical lattice structure weight for these structures is calculated as 25.28 g and 28.16 g respectively, resulting in a theoretical relative density of 10.27 % for the Diamond lattice and 10.43 % for the Kelvin lattice.

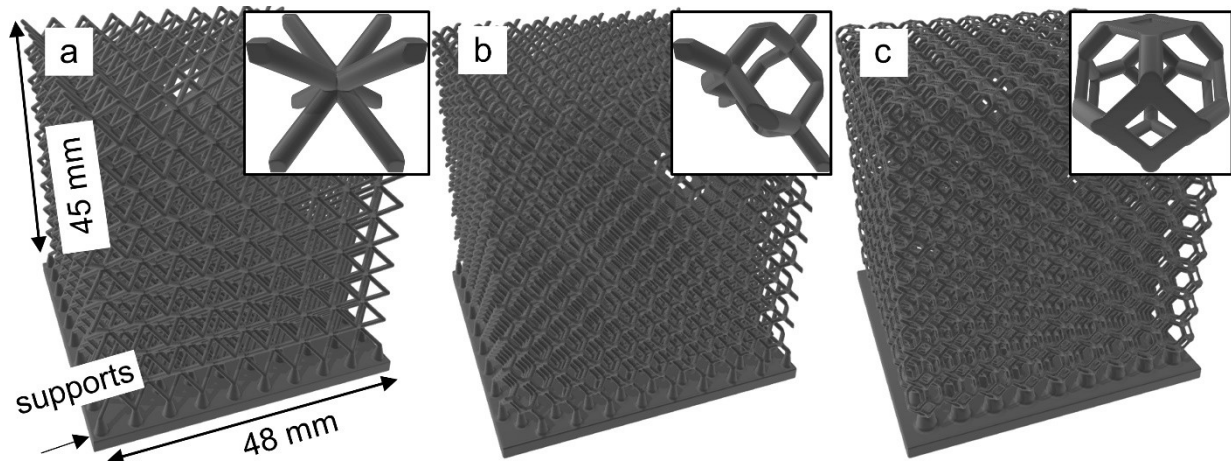


Figure 5-4- Lattice structures within a cube with a length of 45 mm placed on a base plate with supports: a) BCC, b) Diamond, and c) Kelvin.

To establish a consistent hybrid investment casting process, two casting resins from EnvisionTec are initially evaluated using a BCC lattice structure with a strut diameter of 0.73 mm (BCC_0.7) and the AlSi10MgMn alloy. Using the same alloy, the deformation behavior of lattice structures with identical relative densities, but different architectures, including BCC_0.7, Diamond, and Kelvin, is also investigated.

Building on the knowledge gained from earlier process and microstructure studies, the second phase of this section focuses on the combined effects of alloy composition, process parameters, and lattice geometry. For the AlSi10MgMn alloy, BCC structures with varying strut diameters of 0.6 mm (BCC_0.6, relative density of

7.0%) and 0.9 mm (BCC_0.9, relative density of 14.9%) are produced in addition to BCC_0.7. The influence of additional vertical and horizontal struts is also explored for BCC_0.7. Figure 5-5a shows the BCC_0.7 lattice with added horizontal struts (BCCx_0.7), and Figure 5-5b shows the same geometry with additional vertical struts (BCCz_0.7), both with the relative density of 13.2%.

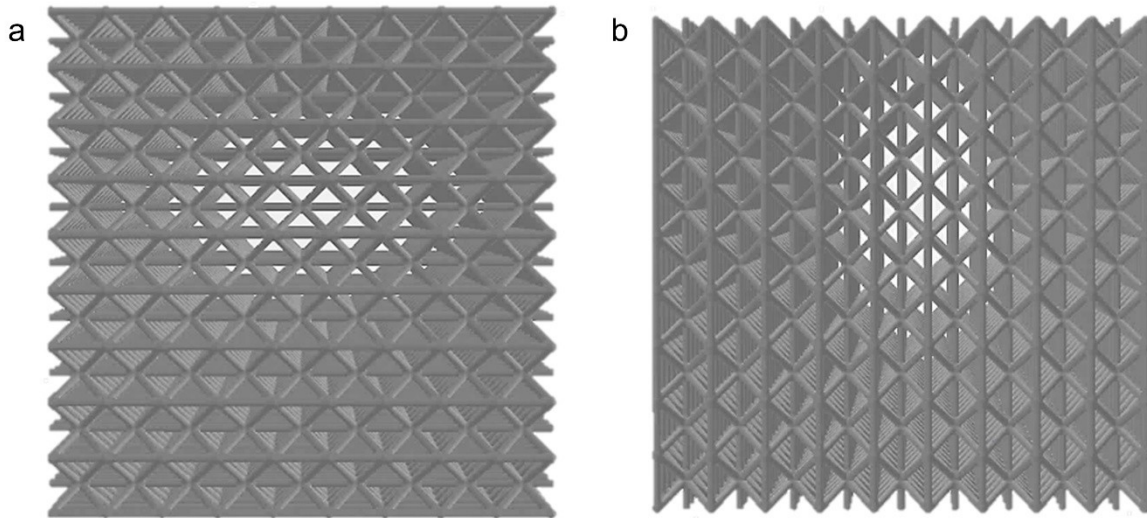


Figure 5-5- BCC lattice structure with a strut diameter of 0.73 mm and with additional struts in: a) the horizontal direction (BCCx_0.7), and b) the vertical direction (BCCz_0.7).

To study the influences of coarse microstructural changes, BCC_0.7 and BCC_0.9 structures are produced using Al-Si alloys with 3, 7, and 10 wt.% Si, both with and without additional struts. In parallel, finer microstructural modifications are introduced through strontium modification and casting in cooler molds to promote a globular grain structure for AlSi7MgMn alloy. These structures are also evaluated in the BCC_0.7 and BCC_0.9 configurations.

Finally, BCC lattice structures with strut diameters of 0.6, 0.7, and 0.9 mm produced by investment casting are compared to their counterparts fabricated via laser powder bed fusion (LPBF), using different printing parameter sets for the AlSi10MgMn alloy. For consistency, the printed samples are designed with the same geometry as the cast BCC structures, having a full size of 45×45×45 mm³. A full list of samples used for each study is provided in the relevant sections.

6. Casting Process

6.1. Block mold casting process window

To obtain a clear understanding of the block molding process window, a type A sample for microstructure examination and a type B sample for mechanical property testing are produced for each casting condition defined in Table 5-1, with the average chemical composition of AlSi7Mg alloy given in Table 4-1. SEM and metallography observations of the struts (Figure 6-1 as an example) were carried out to examine the microstructural characteristics of the foams. The cross-sections exhibited a triangular shape with concave sides, and for foams with a pore density of 10 ppi, the average strut thickness was measured to be $322 \pm 44 \mu\text{m}$ [18]. Ensuring proper mold filling in such fine structures requires careful consideration, as the narrow mold channels make it challenging to distribute the melt completely and evenly.

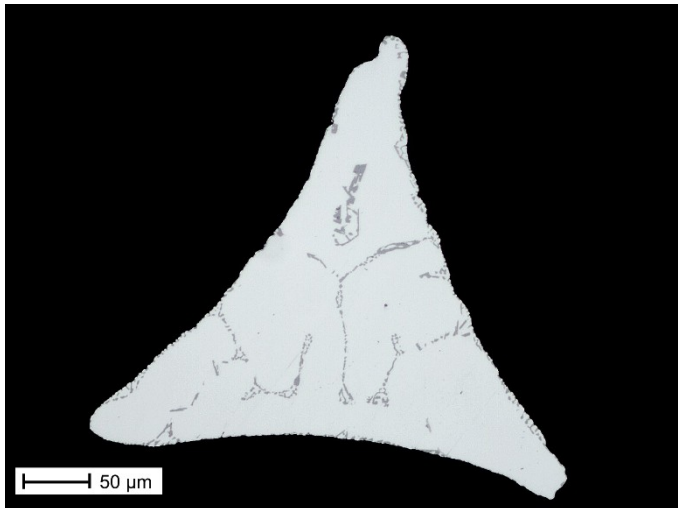


Figure 6-1- Strut cross section of a 10 ppi stochastic foam.

The extent to which a casting metal is able to reproduce the mold contour is referred to as mold filling capacity. If this capacity falls below an acceptable range, the casting edges become rounded, and the mold may not be fully filled. To establish the process window for foam casting, mold filling was assessed by comparing the actual length of the produced samples in each condition to their expected full length. For this evaluation, specimens from type B configuration was used, and mold filling was calculated based on Equation 10.

$$\text{Mold filling (\%)} = \frac{120 \text{ mm} - \text{produced length (mm)}}{120 \text{ mm}} * 100 \quad \text{Equation 10}$$

The average mold filling for all casting conditions are presented in Figure 6-2 for AlSi7Mg alloy. These results are analyzed in relation to the alloy's liquidus and eutectic temperatures, which were determined through thermal analysis of the

melt, giving the mean values of 614.1 °C and 574.0 °C, respectively. The error bars in the mold filling diagrams represent the standard deviation from three samples.

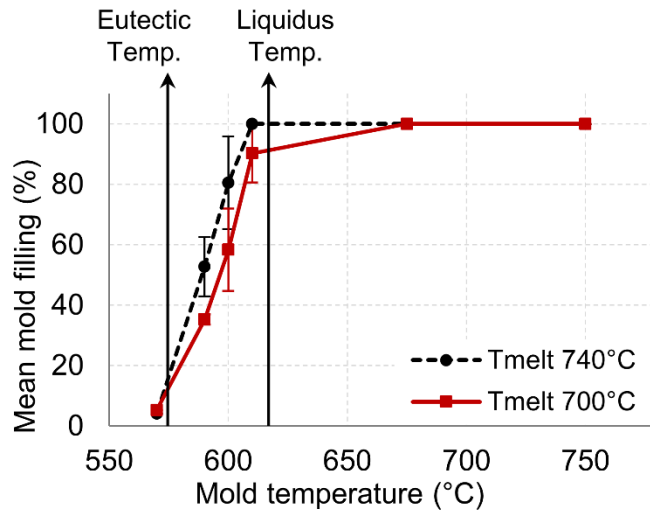


Figure 6-2- Average mold filling percentage at different casting conditions for AlSi7Mg [44].

At all mold temperatures, increasing the melt temperature to 740 °C enhanced mold filling. When the mold temperature exceeded the melt's liquidus temperature, complete mold filling (100%) was achieved. However, as the mold temperature dropped below the liquidus point, the filling percentage decreased linearly. A further decrease of mold temperature around the melt's eutectic temperature led to mold filling rates below 20%, making further mechanical property investigations not possible for these conditions.

To ensure good melt flowability and complete mold filling in the foam casting process, the investment molds are maintained at temperatures above 650 °C during casting [79,139]. The influence of mold temperature on strut shape is evident in the SEM micrographs of Figure 6-3, comparing foams cast at mold temperatures of 750 °C and 575 °C. At the higher mold temperature, the struts exhibit sharp edges, indicating excellent mold filling capacity and precise replication of the PUR foam geometry. In contrast, lower mold temperatures led to premature melt solidification in narrow channels, preventing accurate reproduction of the foam's shape and size. When the mold temperature approaches or drops below the eutectic temperature of the melt, additional forces, such as centrifugal casting [71,140] or high under- or overpressure are required to produce sound foam samples. Alternatively, increasing the melt temperature can enhance mold filling to some extent (Figure 6-2); however, the primary factor governing mold filling capacity remains the mold temperature [79].

Several factors must be considered when optimizing mold temperature, especially for functional components such as heat exchangers or medical implants that incorporate both foam and bulk sections. Casting at a high mold temperature is required for the foam part, with adjustments being made based on strut thickness and part complexity. Since the mechanical properties of metal foams are closely linked to their relative density, maintaining the original structure of the lost pattern is crucial. However, according to a previous research, increase of the mold temperature from 700 °C to 750 °C resulted in a decrease in the mechanical properties [79]. Lowering the mold temperature, which reduces mold filling capacity and leads to rounded edges (Figure 6-3c) is undesirable, as it decreases the cross-section and relative density of the foam. However, excessively high mold temperatures can introduce casting defects and cause nonuniform cooling across different mold sections. For components with thicker struts, reducing the mold temperature can be beneficial, as it allows less shrinkage porosity and a finer microstructure, which are unavoidable during casting under normal conditions.

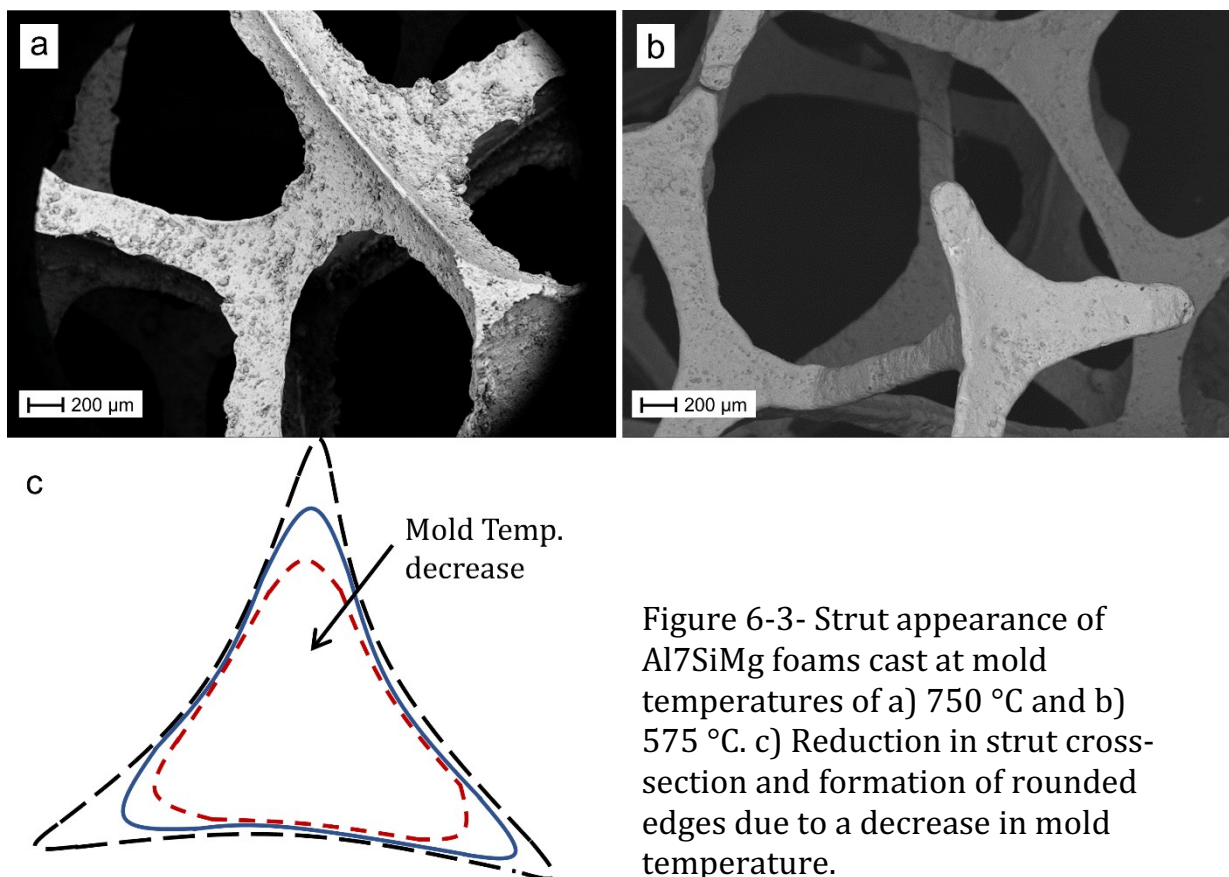


Figure 6-3- Strut appearance of Al7SiMg foams cast at mold temperatures of a) 750 °C and b) 575 °C. c) Reduction in strut cross-section and formation of rounded edges due to a decrease in mold temperature.

6.2. Main features of the foam microstructure

Metallic foams are lightweight structures composed of a metal matrix with voids (pores or bubbles) dispersed throughout. Although, the entire foam structure is recognized as the end product, the metallic part of the foam represents only a

relatively small proportion. The relative density of metal foams, defined as the ratio of foam density to that of the solid metal, typically ranges from 2% to 30%, depending on the material and manufacturing method used.

In the open-cell metallic foams, solidification and microstructure evolution occur within the tiny struts, constrained by limited space. Therefore, microstructure characterization is just focused on the cross-sectional or longitudinal views of the struts. While general microstructural features can be studied based on the principles of bulk materials, the confined solidification space can also introduce unique conditions that influence the final structure.

In aluminum-silicon foams produced under typical casting conditions, where the mold temperature exceeds the liquidus temperature of the melt, aluminum grains solidify in a dendritic form. Solidification is completed through a eutectic transformation, leading to eutectic cells between the dendrite arms. The silicon content directly influences the proportion of silicon particles within the microstructure. Figure 6-4 shows the typical microstructure of Al-Si cast foams with two different silicon contents of 6.9 wt% and 7.8 wt%.

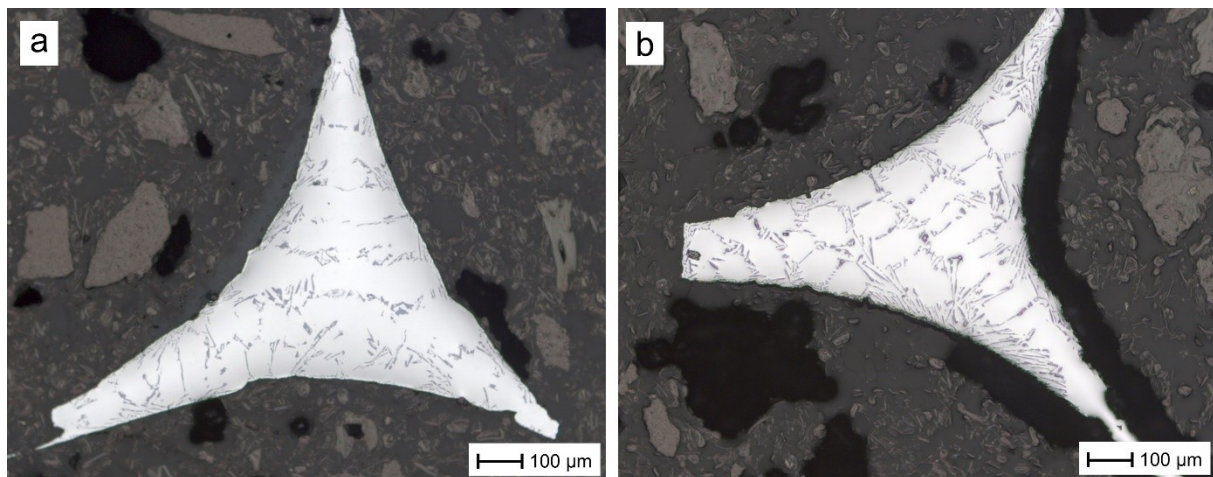


Figure 6-4- Aluminum grains solidified as dendrites under typical casting conditions. Silicon particles form as a result of the eutectic reaction in the Al-Si alloys with (a) 6.9 wt% and (b) 7.8 wt% silicon content.

A detailed analysis of the aluminum grain structure was carried out using anodized metallographic samples examined under polarized light with an optical microscope. This allowed clear visualization of individual aluminum grains in both cross-sectional and longitudinal views of the struts in the Al-Si foams (as in Figure 6-5) produced under the aforementioned casting conditions. In some cases, small aluminum sub-grains, confined to the eutectic silicon regions, were observed along the outer rim of the struts (indicated by white arrows). Abrupt changes in color

within the anodized microstructure signify variations in crystallographic orientation and are indicative of distinct grains. In contrast, gradual or faded color transitions typically do not represent distinct grains but rather result from local segregation of elements and corresponding changes in chemical composition.

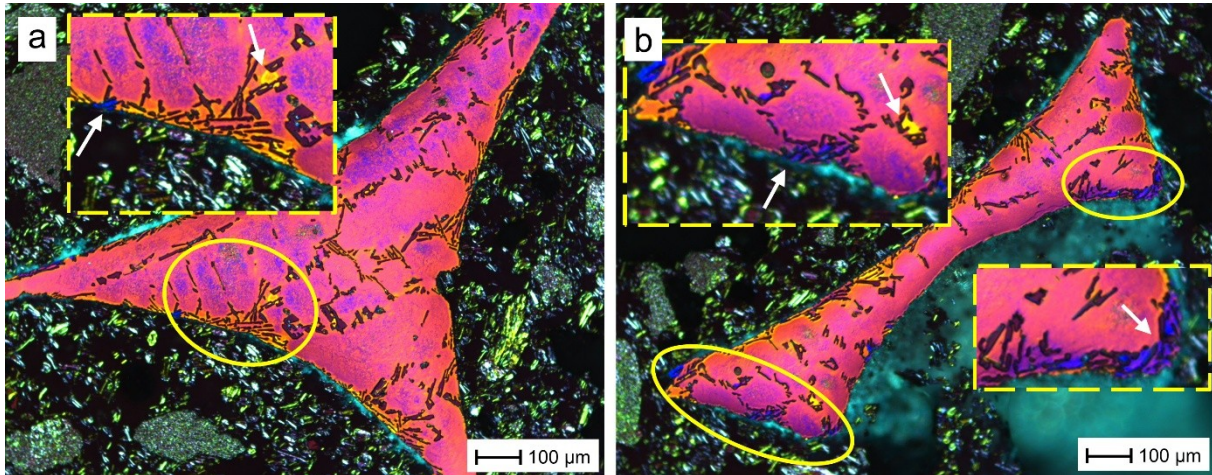


Figure 6-5- Single aluminum grain structure in AlSi7Mg foams produced with mold temperatures of a) 750 °C and b) 675 °C revealed on the anodized samples. Some sub-grains are visible confined to the eutectic silicon regions [44].

Another microstructural feature which is important to study is the presence and type of intermetallic phases. Different intermetallic iron phases can be easily distinguished by metallographic examination based on their distinct coloration, as shown in Figure 6-6. Using EDX analysis, the chemical compositions of these particles were measured. The average chemical compositions, along with the suggested intermetallic components based on literature references [108,113,114] are summarized in Table 6-1.

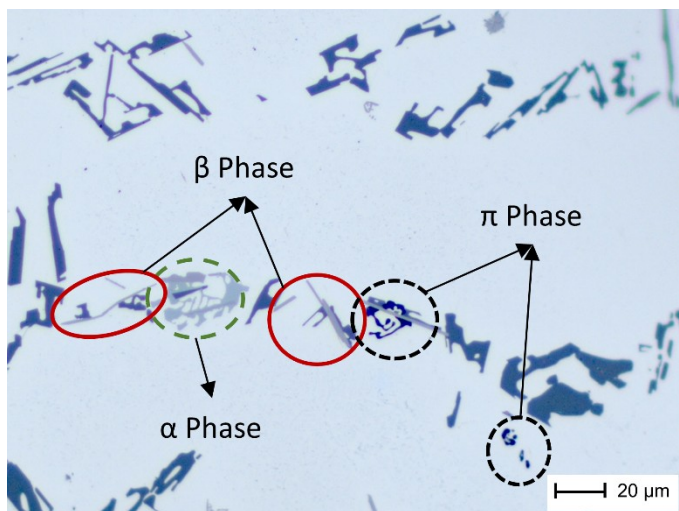


Figure 6-6-Different intermetallic phases in the microstructure of Al-Si cast foams.

Table 6-1- Intermetallic phases as main features of the foam microstructure

Intermetallic phases	Composition/at-%				Suggested compounds	Shape
	Al	Si	Fe	Mg		
α -Fe	~73	~10	~17	-	$\text{Al}_{15}(\text{Fe,Mn})_3\text{Si}_2$ $\text{Al}_{13}(\text{Fe,Mn,Cr,Cu})_3\text{Si}_2$	Chinesse script
β -Fe	~70	~16	~14	-	Al_5FeSi^* $\text{Al}_5(\text{Fe,Mn,Cr,Cu})\text{Si}$	needles, platelet
π -Fe/Mg	~53	~26	~5	~16	$\text{Al}_8\text{FeMg}_3\text{Si}_6$	Chinesse script+dark

6.3. Influence of casting parameters on microstructure

As discussed in section 6.1, mold temperature has a significant impact on the production of an intact and complete cast foam. It influences the microstructure as well.

6.3.1. Aluminum grain structure alteration

Mold temperatures above the liquidus temperature of the melt result in a dendritic aluminum microstructure, as previously shown. By reducing the mold temperature to a range between the liquidus and eutectic temperatures, a gradual transition of aluminum grains from a dendritic to a globular structure was observed (Figure 6-7). By consideration of AlSi7Mg , at a mold temperature of 610 °C, the aluminum structure still exhibited a dendritic morphology with no evident grain boundaries, except for some sub-grain boundaries (as Figure 6-7b). In contrast, at mold temperatures of 600 and 590 °C, the dendritic morphology was substituted by a globular substructure.

The higher number of grains is the result of a higher nucleation potential as the mold temperature decreases. At mold temperatures above the liquidus temperature of the melt, exogenous solidification on the strut surface is not possible. At lower mold temperatures, this is substituted by a high number of nucleation sites on the strut surface and also inside the struts, resulting in a multi-grain structure that no longer resembles the dendritic structure.

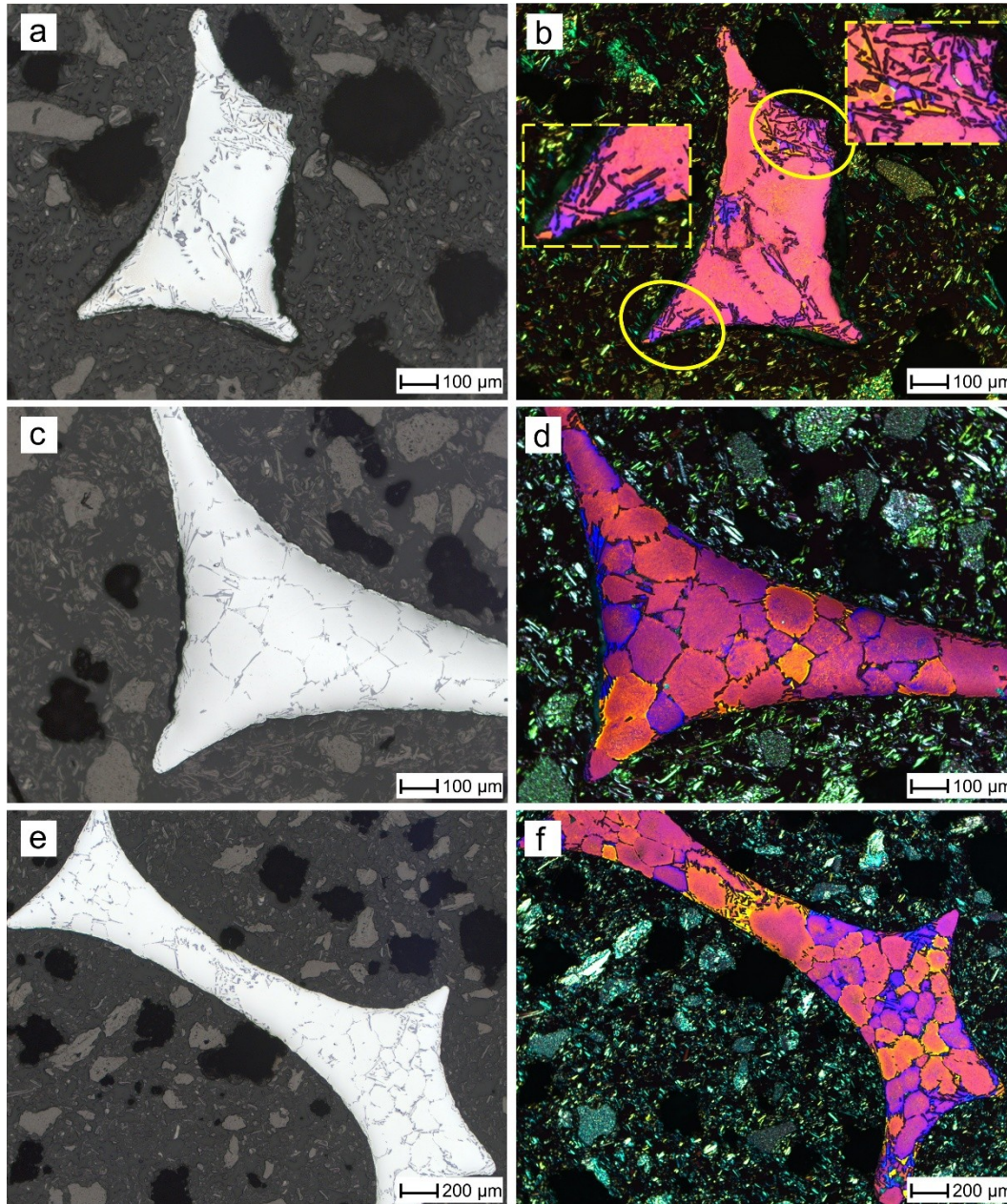


Figure 6-7- Light microscopy micrographs of struts: a) unetched and b) anodized for mold temperature 610 °C, c) unetched and d) anodized for mold temperature 600 °C, e) unetched and f) anodized for mold temperature 590 °C (all with a melt temperature of 700 °C) [44].

6.3.2. Distribution and size alteration of the eutectic silicon particles

A more homogeneous distribution of the eutectic silicon particles is visible in the unetched microstructures shown in Figure 6-7c and e, in comparison to the foams produced at higher mold temperatures. Since eutectic phases form at the end of solidification, they remain between the aluminum grains, but due to the formation of a multi-grain structure, the overall distribution of silicon particles is more homogeneous. At 600 °C and 590 °C, these particles, which form along aluminum

grain boundaries, exhibit a regular alignment and often form quasi-hexagonal outlines.

The spatial distribution of silicon particles was quantified by measuring their mean distance to the nearest strut surface, as shown in Figure 6-8. For this analysis, the distances from each silicon particle to all the surrounding strut interfaces were calculated, and the shortest distance was taken as the particle's effective distance to the surface. At all mold temperatures, silicon particles were found to be slightly closer to the strut surface when the melt temperature was 740 °C compared to 700 °C. Regarding the influence of mold temperature, the eutectic phase tended to form closer to the surface at higher mold temperatures of 675 °C and 750 °C. As the mold temperature decreased, the average distance of silicon particles from the strut surface increased.

The standard deviation in the measured distances reflects the uniformity of silicon particle distribution across the strut cross-section. These values are represented as error bars in Figure 6-8. At the highest mold temperatures, the silicon particles were more concentrated near the strut surfaces, whereas at lower mold temperatures of 590 °C, 600 °C, and 610 °C, the distribution was more scattered and variable.

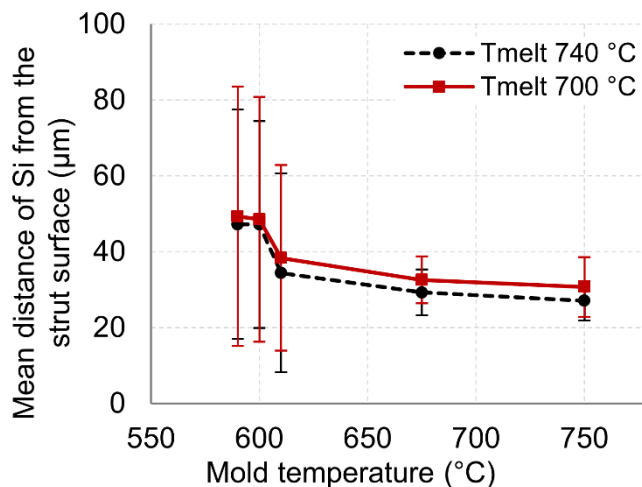


Figure 6-8- Average distance of the eutectic silicon particles from the strut surface at each casting conditions [44].

Figure 6-9a shows the average size of silicon particles under all casting conditions. In general, a melt temperature of 700 °C resulted in larger silicon particles compared to 740 °C across all investigated mold temperatures, which is consistent with previous findings [79]. Interestingly, the trend observed for silicon particle size differ from those typically reported in bulk casting. In conventional bulk casting, increasing the melt and mold temperatures generally leads to slower solidification rates, which promote the growth of larger phases and particles. However, in the current foam casting experiments, a higher melt temperature consistently produced finer silicon particles. Furthermore, the largest average silicon size was observed at

a mold temperature of 675 °C, with smaller sizes measured at both lower and higher mold temperatures. This non-monotonic behavior and the inverse effect of melt temperature emphasize the unique solidification dynamics of foam casting, which differ from what is expected from the usual casting knowledge.

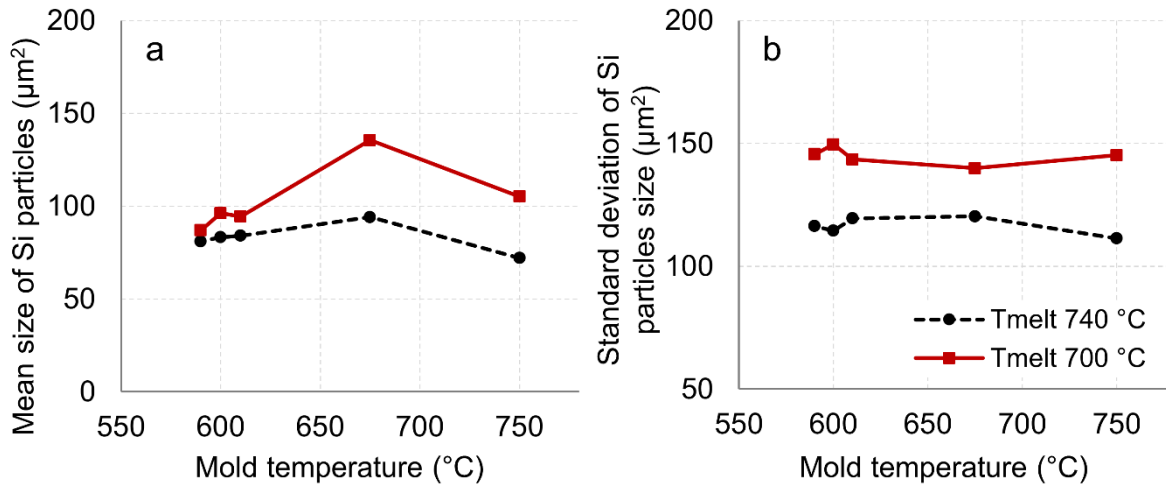


Figure 6-9- a) Mean size and b) standard deviation of silicon particles size at different casting conditions [44].

The silicon particle size distributions were not normal distributions and were skewed toward smaller particle sizes. As a result, using standard deviation as symmetrical error bars misrepresented the actual spread of particle sizes around the mean. Figure 6-9b presents the standard deviation of particle sizes for all casting conditions. It indicates that particle size distribution was more consistent at the higher melt temperature of 740 °C, and that variations in mold temperature had only a minor influence on the standard deviation.

6.3.3. Phases alteration on the strut surface

In Figure 6-10a, micrographs of the surface of a deep-etched sample cast at a mold temperature of 750 °C and a melt temperature of 700 °C reveal that the strut surface was entirely covered by the eutectic phase, with no visible traces of primary aluminum grains. At higher magnification, coarse silicon particles with complex geometries were observed. These particles exhibit a coral-like morphology with three-dimensional growth, characteristic of a eutectic reaction. Their structure appears to radiate outward from a central point, likely corresponding to the original nucleation site.

Reducing the mold temperature affected the distribution of phases on the strut surface. As shown in Figure 6-10b, the deep-etched surface of a sample cast at a mold temperature of 590 °C revealed the presence of aluminum grains between the eutectic regions. In contrast to the higher mold temperature, a smaller portion of

the surface was occupied by eutectic cells. To better understand these microstructural differences across casting conditions, it is essential to consider the progression of solidification within the confined space of the struts.

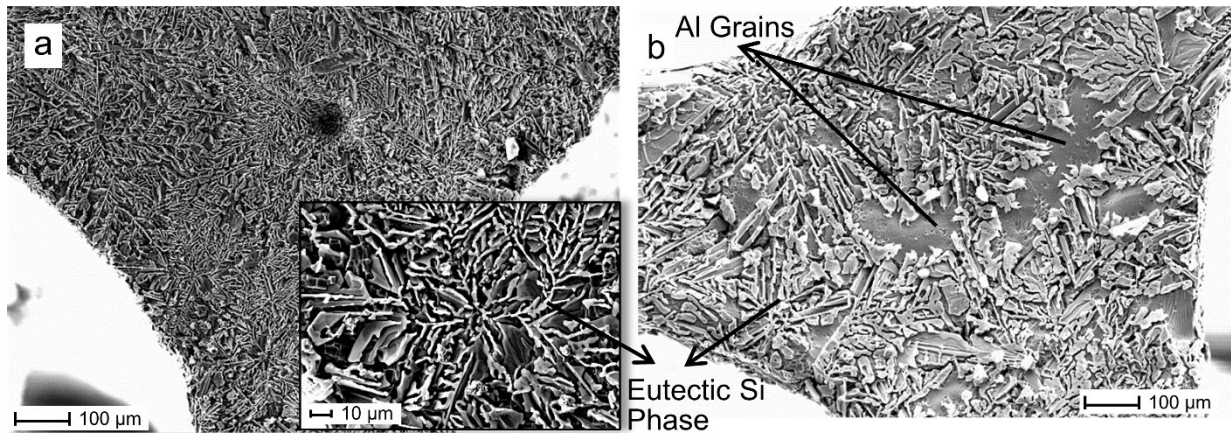


Figure 6-10- SEM micrographs of deep etched foam samples produced at mold temperature of a) 750 °C: a complete coverage of strut surface by eutectic phase and b) 590 °C: primary aluminum grains together with eutectic phase on the strut surface [44].

6.3.4. Microstructure evolution during solidification progress

Molds made from investment material in block mold casting process possess high heat capacity and low heat conductivity. As a result, they maintain high temperature and present slow cooling rates [130]. Consequently, the cooling of cast parts in this process is primarily governed by the thermal behavior of the mold itself. This effect is especially pronounced in thin foam struts, which are entirely surrounded by the hot molding material as their cooling does not occur independently but is controlled by the surrounding mold.

When the mold temperature exceeds the liquidus temperature of the melt, nucleation on the strut surface is not possible until the mold temperature drops below the liquidus temperature. Due to the interconnection of the struts and the cooling effect from convection and radiation at the top surface of the pouring basin, the inner sections of the struts have a higher nucleation potential for the primary aluminum phase. As a result, initial nucleation occurs endogenously within the struts. As the temperature continues to decrease, the nucleated grains grow in a dendritic form toward the mold surface. The formation of a single aluminum grain per strut is attributed to the low nucleation potential and very slow cooling rate. However, as cooling progresses, there is a possibility for the formation of small aluminum sub-grains in the remaining liquid, which are much smaller than the primary aluminum grain and are enclosed by the eutectic phase (as shown in Figure

6-5). These sub-grains probably form near the end of solidification, in areas where local constitutional undercooling occurs.

The endogenous nucleation and the outward solidification process result in accumulation of melt with eutectic composition on the strut surface (Figure 6-10a) and to a lower extent between the dendrite arms. Therefore, in the case of high mold temperatures, the silicon phase is predominantly located near the strut surface [79]. Additionally, because higher melt temperatures increase the overall heat content, the eutectic melt, and consequently the silicon particles, are pushed even closer to the strut surface, as observed for the melt temperature of 740 °C (Figure 6-8).

When the mold temperature during casting is lower than the liquidus temperature of the melt, the nucleation rate increases, both on the mold surface and inside the struts. The high nucleation potential, together with the restricted solidification space in thin struts, limits the dendritic growth of aluminum grains. Accordingly, the single aluminum dendrite per strut is substituted by globular grains for the mold temperatures of 590 °C and 600 °C. However, a microstructure transition from dendritic to globular is observed for a mold temperature just below the liquidus temperature (e.g. 610 °C). A multi grain structure, but still in dendritic form, has also been observed by A.M. Matz et al. in 10 ppi Al7Si foams produced by centrifugal casting in investment molds at 400 °C [140].

The higher number of nuclei under casting conditions 5 to 10 promotes more uniform solidification. Although the silicon particles remain confined between the aluminum grains, their overall distribution across the strut cross-section becomes more homogeneous. As a result, less silicon is visible near the surface, and regions of the primary α -Al phase appear at the strut surface (Figure 6-10b).

6.4. Influence of casting parameters on mechanical properties

6.4.1. Foam deformation behavior

Casting conditions 4, 6 and 8 (mold temperatures of 675 °C, 610 °C and 600 °C, with a melt temperature of 740 °C) were chosen to investigate the influence of dendritic and globular grain structures on the mechanical properties and deformation behavior of AlSi7Mg cast foams. This is illustrated by the DIC maps of the Lagrange strain in the loading direction (ϵ_{xx}) at different strain values (Figure 6-11a) and the mean normalized stress-strain curves (Figure 6-11b). The faded vertical lines, as error bars, showing the standard deviations between different samples from the same casting condition.

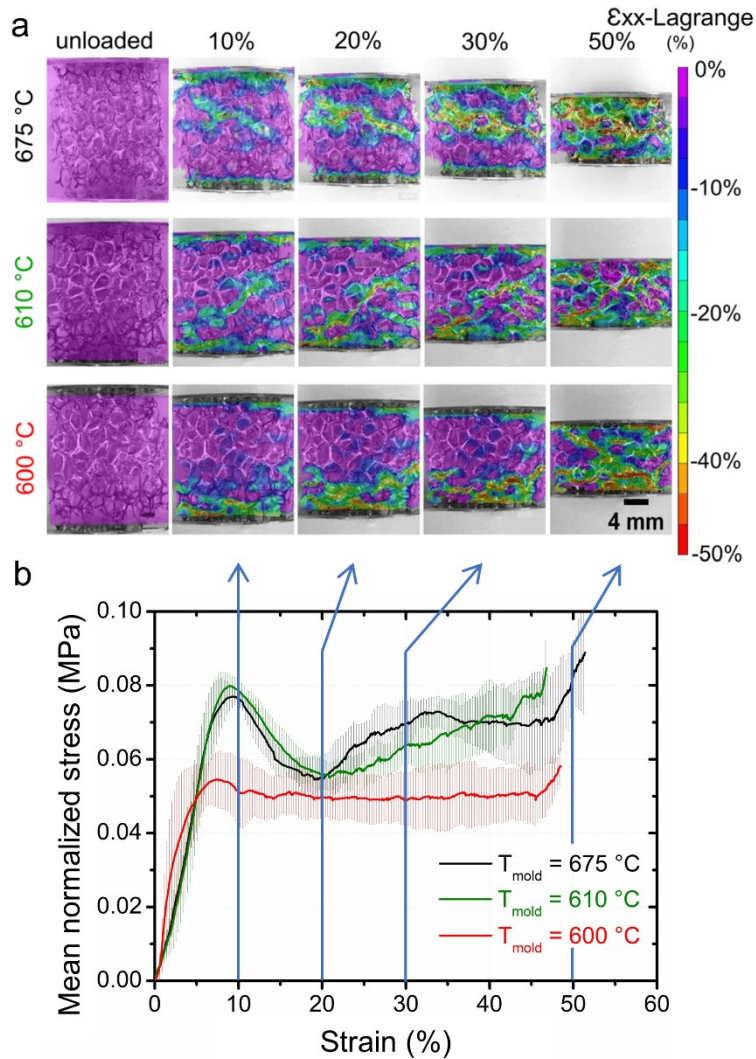


Figure 6-11- Foam deformation and mechanical properties (AlSi7Mg alloy) for different casting conditions: mold temperatures of 675 °C and 610 °C for dendritic grain structure and mold temperature of 600 °C for globular grain structure, a) DIC maps of the Lagrange strain of the loading direction (ϵ_{xx}) and b) normalized stress-strain curves [44,153].

Specimens with a dendritic grain structure and localized silicon particles on the strut surface (mold temperatures of 675 °C and 610 °C) exhibited shear band formation. This initiated shortly after the stress drop following the peak stress (around 10% strain). The shear bands then expanded during subsequent deformation steps, either by widening across the specimen (675 °C) or by forming additional thin bands (610 °C). The normalized stress-strain curves for the dendritic structures displayed sudden stress drops after the peak stress, accompanied by fluctuated plateau regions, which are indicative of the abrupt collapse of the cells.

Specimens with a globular grain structure and a more homogeneous distribution of silicon particles (mold temperature of 600 °C) showed a more uniform layer-by-

layer collapse with only minimal development of shear bands. The first layers in contact with the loading stages on both sides collapsed at around 10% strain, with the thickness of these layers increasing during the later stages of deformation, rather than the formation of shear bands. No distinct stress peak or stress drop was observed for this condition. Additionally, the serration in the plateau region was lower.

Different factors have been identified for the plateau regime serration in foam stress-strain curves. Markaki and Clyne [10] showed that Al-12Si-0.6Mg closed-cell foams with a fine dendritic microstructure exhibited higher strength along with stress oscillations, compared to Al-1Mg-0.6Si with a heterogeneous microstructure and pure Al with coarse dendrites. They attributed the oscillations to brittle failure of the cell walls. Yu et al. [48] demonstrated that larger cell size and fewer cells in closed cell aluminum foams led to an increase in the strength and fluctuations in the plateau regime.

The foam cell size in this study (10 ppi) was relatively large compared to the compression specimen size ($20 \times 20 \times 20 \text{ mm}^3$) extracted from samples type B. Therefore, the microstructure and the deformation behavior of each of the struts become more important when the cell number is limited, which is known as the size effect [159,160]. Scattering of the mechanical properties and overlapping of the error bars can also be explained by the limited number of cells in each specimen. For foams with dendritic microstructure, high peak stresses followed by abrupt stress drops were observed, which is typical behavior for brittle foams. This can be explained by the failure of individual struts and/or the initial collapse of cells [51]. The sudden collapse of cells creates a stress shock to neighboring cells, leading to fracture of more brittle struts. This chain reaction results in serrations in the plateau region of the foam's stress-strain curve.

In a previous study [18], the stress-strain curves of 10 ppi A356.0 foams with a dendritic microstructure and specimen dimensions of $50 \times 50 \times 100 \text{ mm}^3$ did not exhibit large fluctuations. Increasing the specimen size in comparison to the other structural units (e.g. cell size) helps reduce scattering and sudden stress changes, as stable cells that are further away from the original fracture point can maintain the stability of the foam structure [49]. Furthermore, testing larger samples allows for a better observation of deformation characteristics like strain hardening [74]. However, producing larger specimens with the globular microstructure was not successful due to reduction of mold filling when the mold temperature was below the liquidus temperature of the melt. Consequently, for the sake of comparison, all compression tests were conducted on samples with dimensions of $20 \times 20 \times 20 \text{ mm}^3$.

Foams with a globular microstructure, on the other hand, presented flat stress plateaus. The presence of multiple aluminum grains along with the homogeneous dispersion of silicon positively influenced the micro-damage behavior. The dispersion of ductile aluminum grains over the strut cross section enhanced the overall ductility of the struts. Since strut deformation primarily occurs through bending during foam compression, maximum stress is concentrated on the surface of the struts. Therefore, the presence of brittle silicon particles near the strut surface is detrimental to ductility. Previous in-situ deformation investigations conducted inside an SEM [18] observed that brittle fracture initiates from localized silicon particles on the strut surface. A more homogeneous distribution of silicon particles, achieved alongside the globular aluminum grains, clearly enhances the ductile bending behavior of the struts. The size of the silicon particles varied only slightly under the different casting conditions (Figure 6-9) and could not significantly influence the mechanical properties and deformation behavior within the range investigated here.

The normalized stress value was lower for the samples produced with a mold temperature of 600 °C. A reduction in mold filling, caused by the decrease of the mold temperature resulted in thinner struts (Figure 6-3) and lower overall struts stiffness. Therefore, microstructure modifications could not fully compensate the effect of strut thinning on the reduction of tolerable stress. Additionally, a decrease in mold filling also affected the sharpness of the strut edges. Sharp edges increase stress concentration and promote brittle deformation behavior, while rounded edges facilitate smoother deformation. This can be considered another contributing factor to the less brittle deformation observed with the decrease in mold temperature.

6.4.2. Analysis of the strut-level properties

Micro-tensile testing of individual struts provides valuable insight into mechanical behavior at the strut level. However, selecting suitable struts with sufficient length and high surface quality for accurate strain evaluation using image analysis software remains a significant challenge. To address this, ten random single struts were extracted from a foam specimen out of type B sample, and the five with the best geometry and surface condition were selected for micro-tensile testing. The stress-strain responses of these five selected struts from foams cast under different conditions (mold temperatures of 675 °C, 610 °C, and 600 °C, all with a melt temperature of 740 °C; corresponding to casting conditions 4, 6, and 8) are presented in Figure 6-12. The numbering of the individual struts reflects the order in which they were extracted and does not indicate their position within the foam structure. The average values of UTS, uniform deformation strain, and fracture

strain of the individual struts, together with their standard deviations for 5 struts per condition, are reported in Table 6-2.

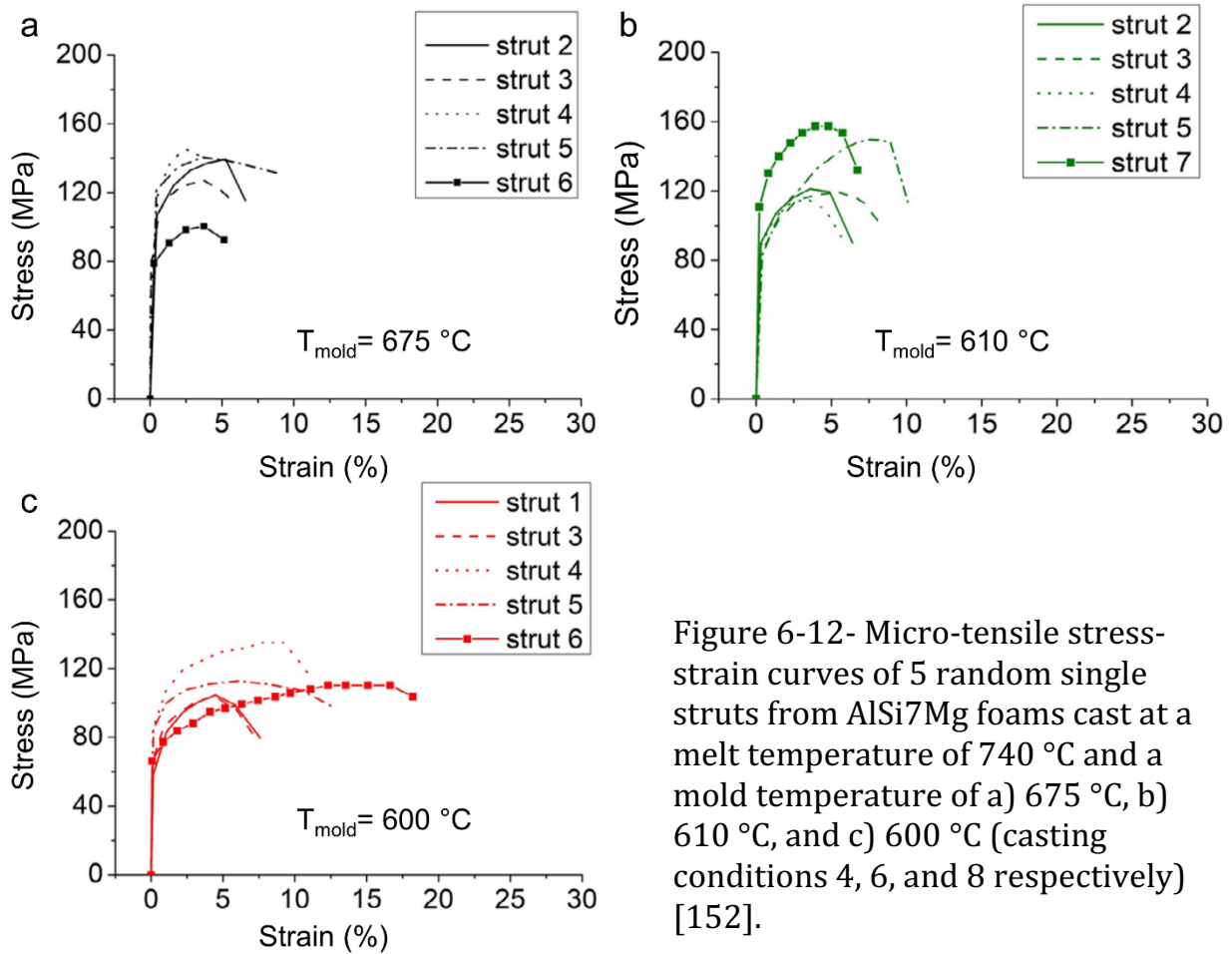


Figure 6-12- Micro-tensile stress-strain curves of 5 random single struts from AlSi7Mg foams cast at a melt temperature of 740 °C and a mold temperature of a) 675 °C, b) 610 °C, and c) 600 °C (casting conditions 4, 6, and 8 respectively) [152].

Table 6-2- Average values of UTS, uniform deformation strain, and fracture strain of 5 single struts tested in accordance with Figure 6-12 (standard deviations are shown in parentheses) [153]

Melt temperature	740 °C		
Mold temperature	675 °C	610 °C	600 °C
Grain structure	Dendritic	Transition (close to dendritic)	Globular
UTS (MPa)	130.7 (16.1)	132.7 (17.7)	114.0 (11.1)
Uniform deformation strain (%)	3.7 (0.9)	4.6 (1.3)	8.0 (4.4)
Fracture strain (%)	6.0 (1.7)	7.3 (1.6)	11.2 (4.0)

While the globular grain structure exhibits lower strength at the strut level, its average uniform elongation is nearly double that of the dendritic microstructure. This observation supports the earlier discussion on foam deformation behavior, confirming that a polycrystalline, globular grain structure, combined with a more homogeneous distribution of silicon particles, enhances the ductility of the struts. In contrast, the localized presence of the strong yet brittle eutectic silicon phase on the strut surface contributes to higher strength but also increased brittleness.

Fractographic analysis of the struts following micro-tensile testing revealed a rough fracture topography associated with the dendritic grain structure. The irregular orientation of eutectic Si particles positioned between the dendrite arms significantly influences the fracture morphology, leading to a jagged and uneven fracture surface. Representative examples of such fracture surfaces are shown in Figure 6-13a and b for the mold temperatures of 675 °C and 610 °C (strut 4 and strut 7, respectively).

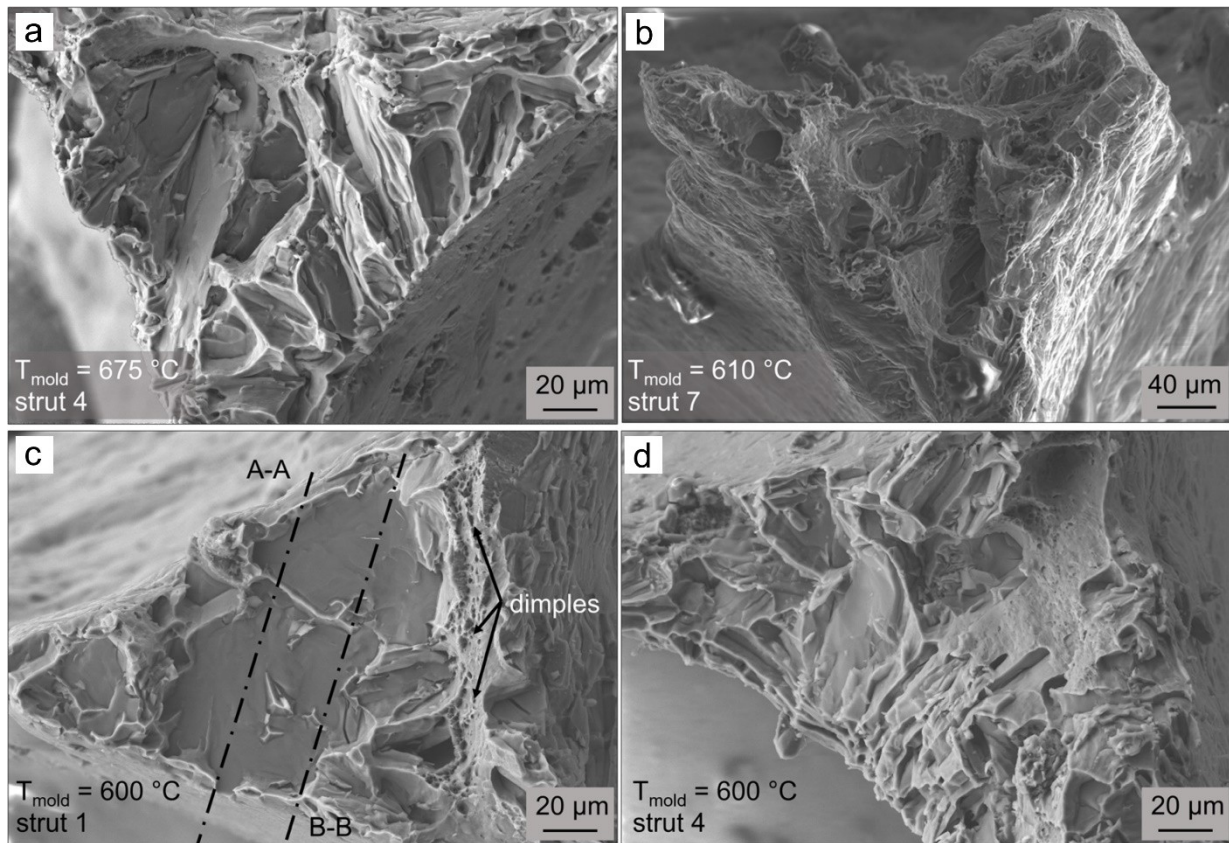


Figure 6-13- Fracture surface analysis by SEM after micro-tensile test of struts: a) strut 4 from the mold temperature of 675 °C, b) strut 7 from the mold temperature of 610 °C, c) strut 1 and d) strut 4 from the mold temperature of 600 °C [153].

The struts with a globular grain structure (casting condition 8) exhibited a higher standard deviation in both uniform deformation and fracture strains, indicating greater variability in mechanical performance compared to those with a dendritic

structure. To investigate the cause of this scatter, two struts with markedly different strengths and fracture strains were selected for fracture surface analysis (Figure 6-13c and d). The examination revealed two distinct fracture topographies. A rough fracture surface (like strut 4) is indicative of transgranular fracture, where the strut exhibited higher strength. In contrast, the flat fracture surface of strut 1 suggests intergranular failure along silicon plates. Elemental mapping by EDX across lines A-A and B-B confirmed the presence of coarse silicon particles on the flat fracture regions, while fine dimples along the strut's side were associated with aluminum.

In this microstructure, the silicon particles are aligned along the grain boundaries, making these interfaces inherently brittle. When fine Si particles are oriented parallel to the loading direction, they can impede dislocation motion, thereby increasing the strut's strength. This condition is typically associated with a rough fracture surface, similar to that observed in dendritic structures. Conversely, when the Si particles are coarse and oriented perpendicular to the load direction, dislocation movement is less effectively hindered. In such cases, the weak interface between the silicon particles and the aluminum grains is more likely to separate, leading to premature failure and a flat fracture surface. Consequently, the variation in mechanical properties at the strut level can be attributed to microstructural inhomogeneity within the thin struts. Importantly, this variation is often governed by the local microstructure in the weakest region of the strut where fracture initiates, rather than by the minimum cross-sectional area alone.

6.5. Nonuniformity in block mold casting

6.5.1. Cooling nonuniformity in the longitudinal direction

Temperature measurements on long foam samples (type C) during casting and solidification were conducted to understand the cooling conditions in the longitudinal direction. The results for commercially pure aluminum and Al7SiMg are presented in Figure 6-14. Since the thermocouple measured the temperature of the molding material surrounding the foam struts at the thermocouple's location, it was not possible to directly track the characteristic solidification temperatures. In most cases, the temperature profile showed continuous cooling controlled by the investment material. However, in both alloys, the first section was strongly influenced by the melt in the pouring basin, and the phase transformations of the significant amount of melt in the basin was also detectable in the foam sections close to it.

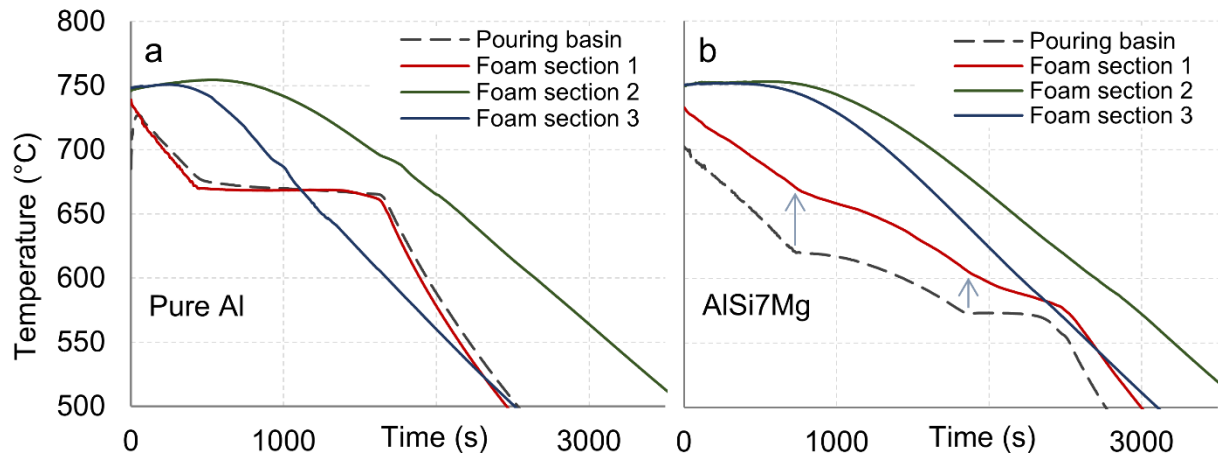


Figure 6-14- Foam cooling at different sections of sample type C for a) a commercially pure Al and b) an AlSi7Mg alloy.

The high latent heat of solidification of pure aluminum significantly altered the cooling behavior in section 1, as shown in Figure 6-14a resulting in a temperature profile similar to that of the pouring basin. In the case of the alloy, deviations in the cooling curve at the time of liquidus and solidus transformations were observed in the same section, as indicated by arrows in Figure 6-14b. Section 2, located in the middle of the mold, showed a lower cooling rate compared to section 3, located at the bottom of the mold, in both materials. Since the primary microstructural features develop during solidification, the cooling rates were calculated in the temperature range between 620 °C and 570 °C (approximate liquidus and solidus temperatures of the AlSi7Mg alloy) and are reported in Table 6-3.

Table 6-3- Cooling rates of different foam sections in sample type C in the temperature range of 620-570 °C

Foam section	Cooling rate (°C/s)	
	Commercially pure Al	AlSi7Mg
1	No linear cooling	0.05
2	0.09	0.09
3	0.12	0.11

Although the overall cooling rates are relatively low in this process compared to other casting processes, the cooling conditions across different sections of the mold were notably inconsistent. The pouring basin acts as a heat source, reducing the cooling rate in the nearby regions. Moving away from the basin, the cooling rate increases, with the highest rate observed at the opposite end of the foam, where less mold material surrounds the structure. In this area, the cooling rate is nearly twice that of section 1. To achieve more uniform cooling along the longitudinal axis,

additional investment material should be poured above the foam during mold preparation to slow down the rapid cooling in section 3. This adjustment was implemented in the production of all specimens intended for mechanical properties testing to ensure better consistency (including samples type A and B produced at different casting conditions, samples type D, E, and B-Plus to study the effect of second phases, and all of the lattice structures, excluding samples type C to study the effect of cooling nonuniformity in longitudinal direction).

6.5.2. Microstructure alteration in the longitudinal direction

Type C samples made from commercially pure aluminum, AlSi7Mg, and AlSi7MgSr400 alloys were cast to investigate the influence of nonuniform cooling on microstructure development. The results revealed morphological variations in the iron intermetallic phases along the length of the samples. In the longitudinal section 1, where the cooling rate was lower, the iron intermetallic predominantly appeared as β -phase platelets. In contrast, at longitudinal section 3, where the cooling rate was higher, more compact Chinese script morphologies were formed, as shown in Figure 6-15.

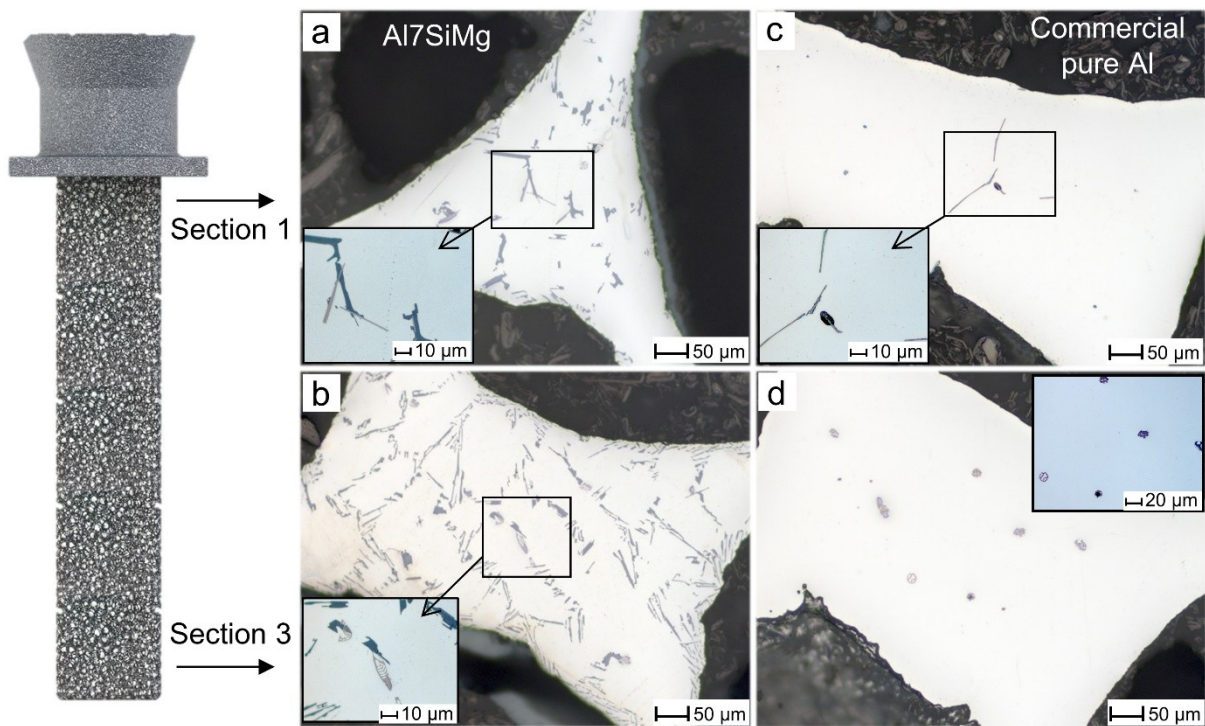


Figure 6-15- Microstructural deviation, in term of iron intermetallic phase, at different foam sections of sample type C due to the nonuniformity in cooling. For AlSi7Mg: a) β -iron phase at section 1 and b) Chinese script phases at section 3. For commercially pure aluminum: c) β -iron phase at section 1 and d) globular phase at section 3.

Practical experiments have shown that the formation of the β -phase in bulk casting can be suppressed by increasing the cooling rate. This alteration affects the phase formation temperature and promotes the retention of the α -phase by increasing its formation temperature and reducing the time available for the growth of needle shape β -phase [109,161]. However, the variations in cooling rate observed in this study were not significant enough to explain the observed microstructural differences alone. Instead, these differences are likely influenced by the unique geometry and localized solidification dynamics of foam casting, which must be considered in addition to thermal effects.

In foam casting, where strut thicknesses are in the micrometer range, diffusion dynamics, solidification behaviors, and heat transfer mechanisms differ markedly from those in bulk casting. The hot investment material surrounding these thin melt zones together with the high surface area-to-volume ratio of foam struts play a dominant role in governing the thermal profile of the solidification area. Under such conditions, even modest increases in cooling rates—from 0.05 to just over 0.1 °C/s—can shift the phase transformation behavior. Specifically, the β -phase may transform into the Chinese script phases (such as α -phase or π -phase in the presence of magnesium), due to the limited time available for diffusion-controlled growth of the β -phase into thick needle-shape particles.

The higher sensitivity to microstructural changes is pronounced in foam casting, where the combination of controlled cooling rate and specific strut geometry provides a favorable platform for fine-tuning the microstructure. This behavior was also observed in the modification of the silicon phase by Sr addition; in Section 3, a refinement of the silicon particles in the AlSi7MgSr400 alloy was successfully achieved, while in Section 1 no such improvement was observed (Figure 6-16).

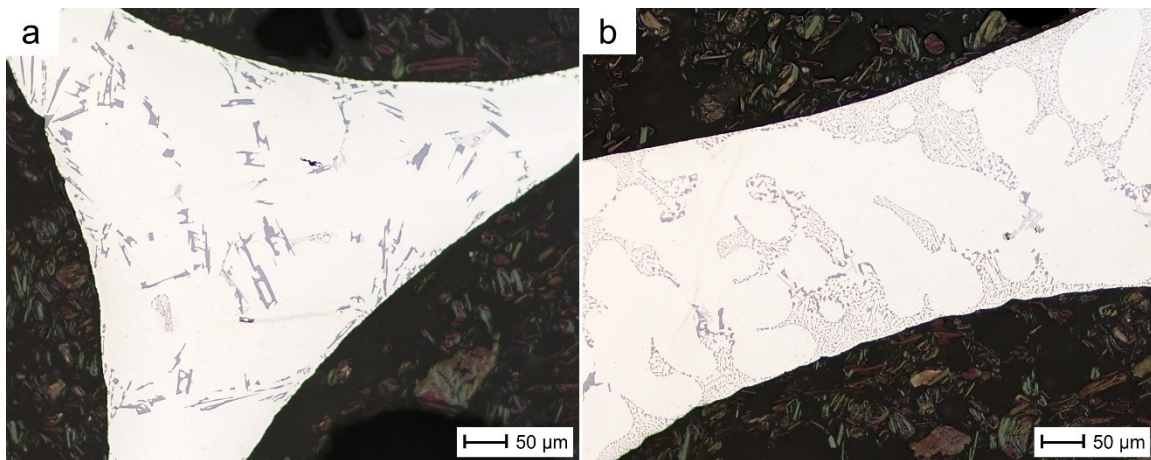


Figure 6-16- Microstructure of AlSi7MgSr400 alloy in type C sample in a) section 1 and b) section 3.

As explained, the β -iron phase is the most undesirable intermetallic phase in aluminum alloys, and efforts should be made to minimize its formation through optimization of the chemical composition or casting process. However, complete avoidance of this phase is not feasible overall due to the inherent cooling inhomogeneity in the process. As a result, the addition of manganese will be explored in the next phase of the research as a potential solution to suppress the formation of this intermetallic phase.

6.5.3. Cooling nonuniformity in the horizontal direction

Temperature measurements were also conducted on radial foam samples (type D as Figure 5-2) at three zones to assess the cooling rates during solidification and evaluate the horizontal cooling consistency, as illustrated in Figure 6-17.

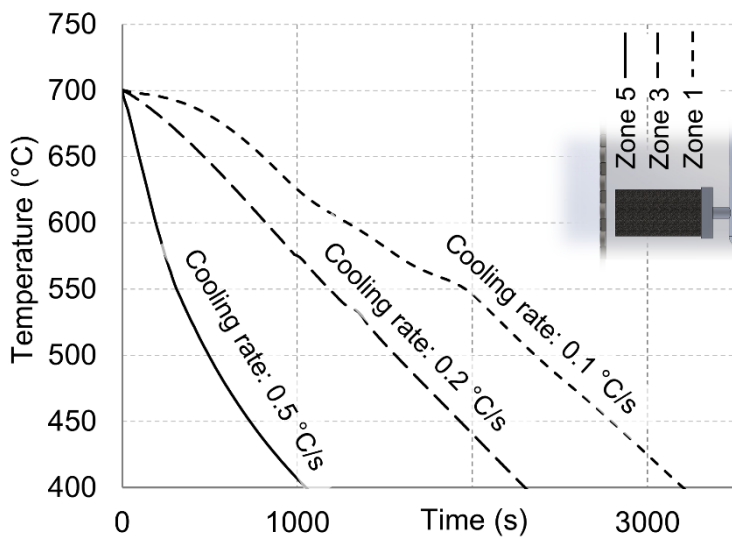


Figure 6-17- Cooling curves and cooling rates for zones 1, 3, and 5 of the radial sample type D in the horizontal direction of flaks [45].

As already mentioned, since the struts are entirely covered by the molding compound, the temperature measurements are representative of the foam cooling at the position of the thermocouples. The use of hot investment molds for complete filling of the ultra-fine structure leads to slow cooling rates, with the amount of surrounding molding material determining the cooling conditions. Consequently, foam sections closer to the flask wall experience locally elevated cooling rates due to the reduced surrounding molding material and increased heat transfer from the cuvette surface. As a result, the outer sections of the radial samples cooled approximately five times faster than the center of the mold. The high proportion of molding material along the longitudinal direction of the molds created a lower thermal gradient in that direction, making the radial direction the primary heat transfer pathway for these samples.

6.6. Conclusion: Process–microstructure link

In the production of cast open-cell aluminum foams, several critical process-related aspects must be considered:

- Mold temperature plays a crucial role in achieving accurate geometric replication of the foam, as dictated by the PUR pattern design.
- When applying a medium vacuum to the casting flask during casting, mold temperatures above the liquidus point of the melt are essential to ensure complete filling and accurate reproduction of the intended geometry.
- During cooling and solidification, non-uniform thermal behavior is observed both in longitudinal and radial directions. In the longitudinal direction, the cooling rate at the foam end is up to twice higher as at the section close to the pouring basin. In the radial direction, the area adjacent to the flask wall cools up to five times faster than the central zone.
- These non-uniform cooling rates lead to microstructural variations, particularly due to changes in local solidification conditions. Although the differences in cooling rate are moderate, their effects on microstructure are noticeable. This highlights the high sensitivity of foam casting to process-induced variations, owing to the combination of thin strut geometries and moderately slow cooling, which changes the local solidification kinetics and provides an opportunity for microstructural fine-tuning.

Regarding the effects of the general microstructure occurrence on the properties:

- A high mold temperature leads to endogenous nucleation and dendritic solidification toward the strut surface, resulting in an accumulation of silicon particles on the surface and promoting brittle fracture of the struts during deformation.
- A mold temperature between the liquidus and eutectic temperatures leads to a globular multi-grain structure of aluminum and a more homogeneous distribution of silicon particles, which increases ductility at both the foam and strut levels compared to the dendritic microstructure.

7. Microstructure Adjustment

Solidification primarily determines the overall morphology of aluminum grains and the eutectic phase, which represent the coarse features of the microstructure. In contrast, finer aspects can be tailored through targeted alloy design. One of the most effective fine-tuning strategies for Al-Si alloys is silicon modification, which is particularly beneficial for foam casting that solidify under low cooling rates and tend to form coarse silicon particles. When the goal is to generate finer features on the micro- and nanoscale as well, adopting a heat-treatable alloy concept allows for the in-situ formation of precipitates with controlled size.

7.1. Chemical modifier

Various modifying elements, including sodium, strontium, and europium, have been investigated for AlSi7Mg under different cooling rates (in accordance to Table 5-3) to identify the most effective approach for modifying cast Al-Si open-cell foams. Although sodium is known for its strong ability to refine eutectic silicon, its high reactivity, risk of gas entrapment, and difficulty in maintaining precise concentration levels in the melt presented significant challenges in the laboratory scale. As a result, its usage in this study was limited to one test casting to evaluate its feasibility for aluminum foams. Strontium, on the other hand, has proven to be a more practical alternative, providing similar modification effects with easier handling and improved process stability. Europium was also investigated to determine whether it might offer superior performance under low cooling rates observed in foam casting [162,163]. The modification degree (Equation 9) by adding 200, 400 and 800 ppm Sr and Eu to AlSi7Mg at different cooling rates is shown in Figure 7-1.

Table 7-1- Chemical composition of modified AlSi7Mg melts cast into different molds (as in Table 5-3) to study the modification degree at various cooling rates

Melt ID	Si (wt%)	Fe (wt%)	Ti (wt%)	Sr (ppm)	Eu (ppm)	Melt temp. (°C)
200 Sr	6.95	0.08	0.12	212	---	705
400 Sr	6.98	0.07	0.11	423	---	703
800 Sr	7.05	0.07	0.11	851	---	705
200 Eu	6.89	0.07	0.12	---	233	701
400 Eu	7.01	0.09	0.11	---	398	705
800 Eu	6.93	0.08	0.11	---	812	704

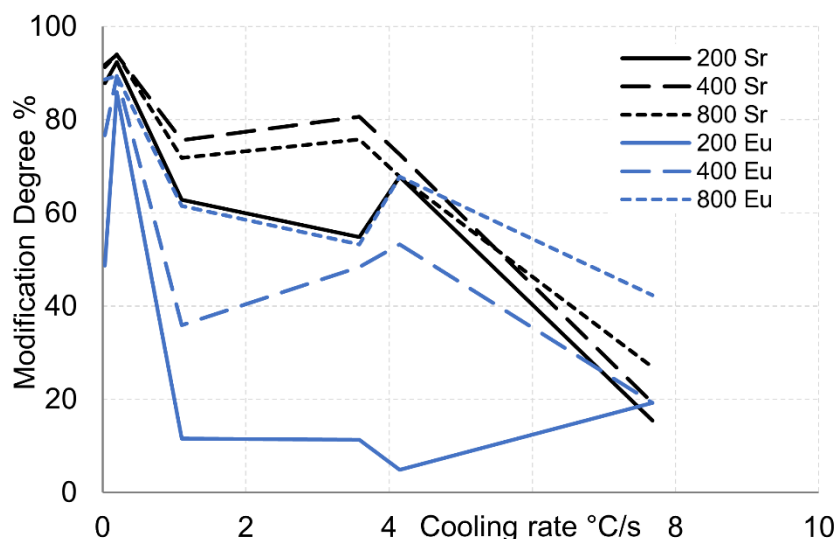


Figure 7-1-
Modification degree
for AlSi7Mg at
different cooling rates
by Sr and Eu at
different contents.

A general trend was observed showing that at higher cooling rates, the effectiveness of chemical modifiers is reduced. This is because rapid solidification inherently leads to finer silicon particles, diminishing the influence of modification elements in further refining silicon size. In contrast, at slower cooling rates, which are particularly relevant for aluminum foam casting, modifiers have a more pronounced effect on reducing silicon particle size. Despite this overall trend, some intermediate data points showed deviation due to scatter in the silicon size measurements and variability in the microstructural analysis.

As can be seen from Figure 7-1, strontium consistently induced a higher degree of silicon modification across nearly all cooling rates. While both 400 ppm and 800 ppm Sr additions led to similar modification levels, the 400-ppm content demonstrates a more favorable and uniform modification effect.

This could be attributed to the saturation behavior of strontium in the melt. At 400 ppm, the amount of Sr may be sufficient to modify the eutectic silicon effectively without introducing side effects such as Sr-rich intermetallics or poisoning effects from excessive Sr. In contrast, at 800 ppm, the excess Sr may lead to local over-modification, coarse intermetallic formation, or uneven distribution, slightly reducing the overall effectiveness. Therefore, by consideration of this pre-test on different cooling rates, 400 ppm appears to be a more efficient and balanced concentration for achieving consistent modification.

Europium achieved a modification level comparable to that of strontium only at the highest tested concentration (800 ppm Eu). However, due to its significantly higher cost and the relatively large quantity required to produce a notable effect, europium is not considered a practical choice for this application. The lower efficiency of Eu at moderate concentrations could be attributed to its lower affinity for segregation

at the solid–liquid interface compared to strontium, which may limit its effectiveness in modifying the silicon eutectic under typical casting conditions [99]. Consequently, strontium was chosen for the further investigation of the silicon phase modification in Al-Si foams.

7.1.1. Modification by strontium

To evaluate the potential for silicon modification within the fine strut structures and to investigate the influence of cooling rate, radial foam samples (type D) and mechanical properties specimens (type B-Plus) were cast using AlSi7Mg and AlSi10Mg alloys with varying amounts of strontium (see Table 7-2). Based on the cooling results from section 6.5.3, each sample was divided into five cooling severities. Classification of the cooling severities was directly related to the zone division previously used for the temperature measurements in Figure 6-17. Zone 5, located closest to the flask wall, exhibited the highest cooling rate of 0.5 °C/s, while Zone 1, at the center of the mold, experienced the slowest cooling rate of 0.1 °C/s. The metallographic study revealed apparent changes due to the variation in cooling for both modified and unmodified conditions, as shown in the micrographs of Figure 7-2. The cross-section of a metallographic sample is shown in Figure 7-2a to illustrate how a sample type D is divided into 5 separate zones corresponding to different cooling rates.

Table 7-2- Chemical composition and casting condition of Al-7Si and Al-10Si melts with and without Sr addition for production of type D and B-Plus specimens

Melt ID	Si (wt%)	Fe (wt%)	Mg (wt%)	Ti (wt%)	Sr (ppm)	Melt temp. (°C)	Mold temp. (°C)
Al-7Si	6.92	0.09	0.29	0.12	--	702	700
Al-7Si-200Sr	6.96	0.08	0.29	0.12	214	703	
Al-7Si-400Sr	7.02	0.08	0.29	0.11	436	706	
Al-7Si-800Sr	6.99	0.08	0.29	0.11	826	703	
Al-10Si	10.43	0.12	0.25	0.09	--	682	
Al-10Si-200Sr	10.35	0.10	0.24	0.08	237	682	
Al-10Si-400Sr	10.53	0.10	0.24	0.09	438	685	
Al-10Si-800Sr	10.38	0.11	0.25	0.10	804	683	

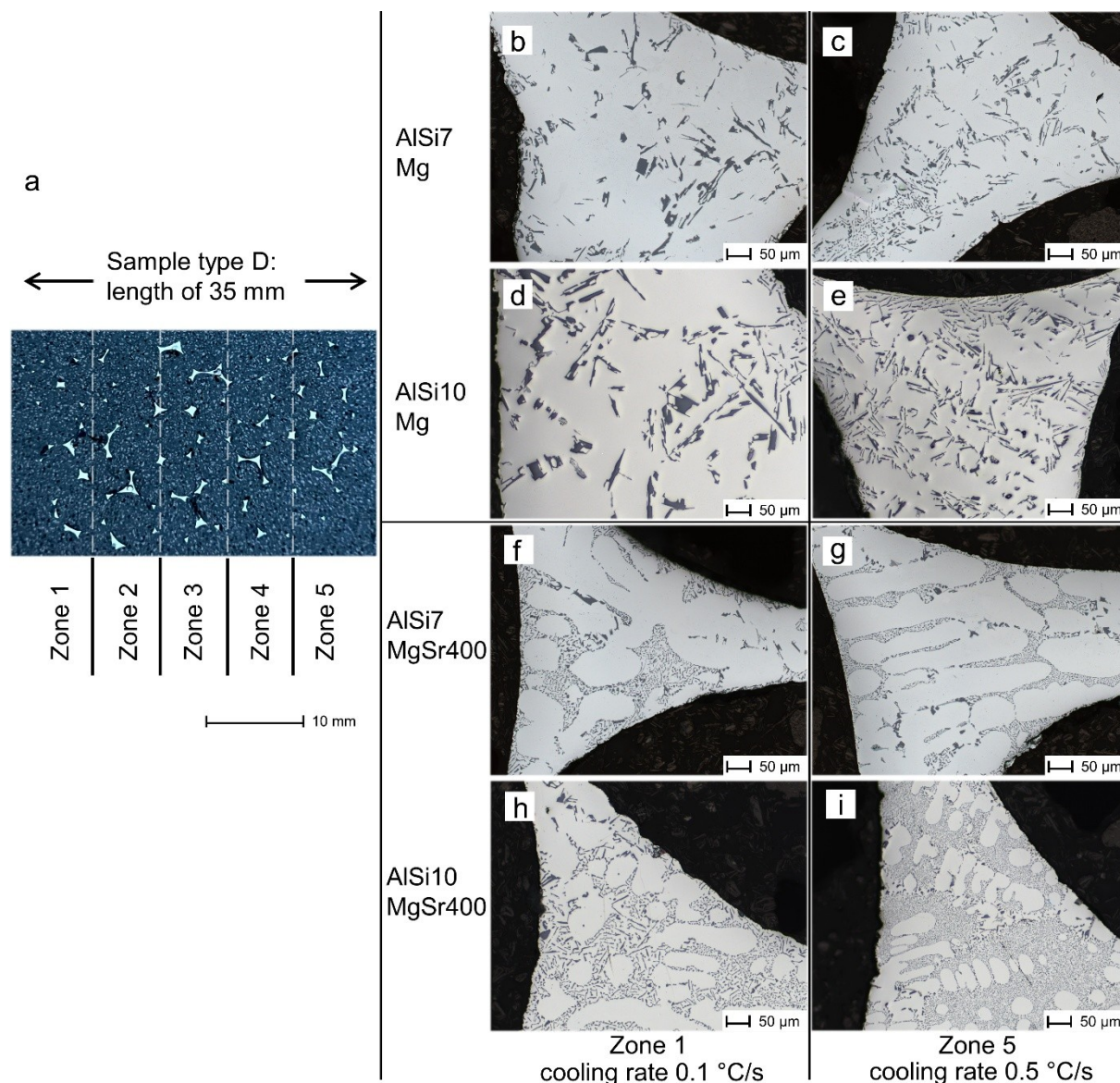


Figure 7-2- Microstructure study of radial samples, a) division of the longitudinal section of a type D sample into five zones. Microstructure of Al-7Si at b) zone 1 (the least cooling severity) and c) zone 5 (the highest cooling severity); microstructure of Al-10Si at d) zone 1 and e) zone 5; microstructure of Al-7Si-400Sr at f) zone 1 and g) zone 5; microstructure of Al-10Si-400Sr at h) zone 1 and i) zone 5 [45].

An analysis of the silicon particle characteristics revealed information on the average size and roundness of the silicon particles in accordance to the cooling rate (Figure 7-3). For all the alloys, Si particles became smaller and more uniform in size at higher cooling rate. The cooling intensity had no significant influence on the roundness of the Si particles, however, the addition of Sr resulted in rounder particles. Moreover, strontium was more effective in refining the silicon phase in the alloy with a higher silicon content (AlSi10Mg).

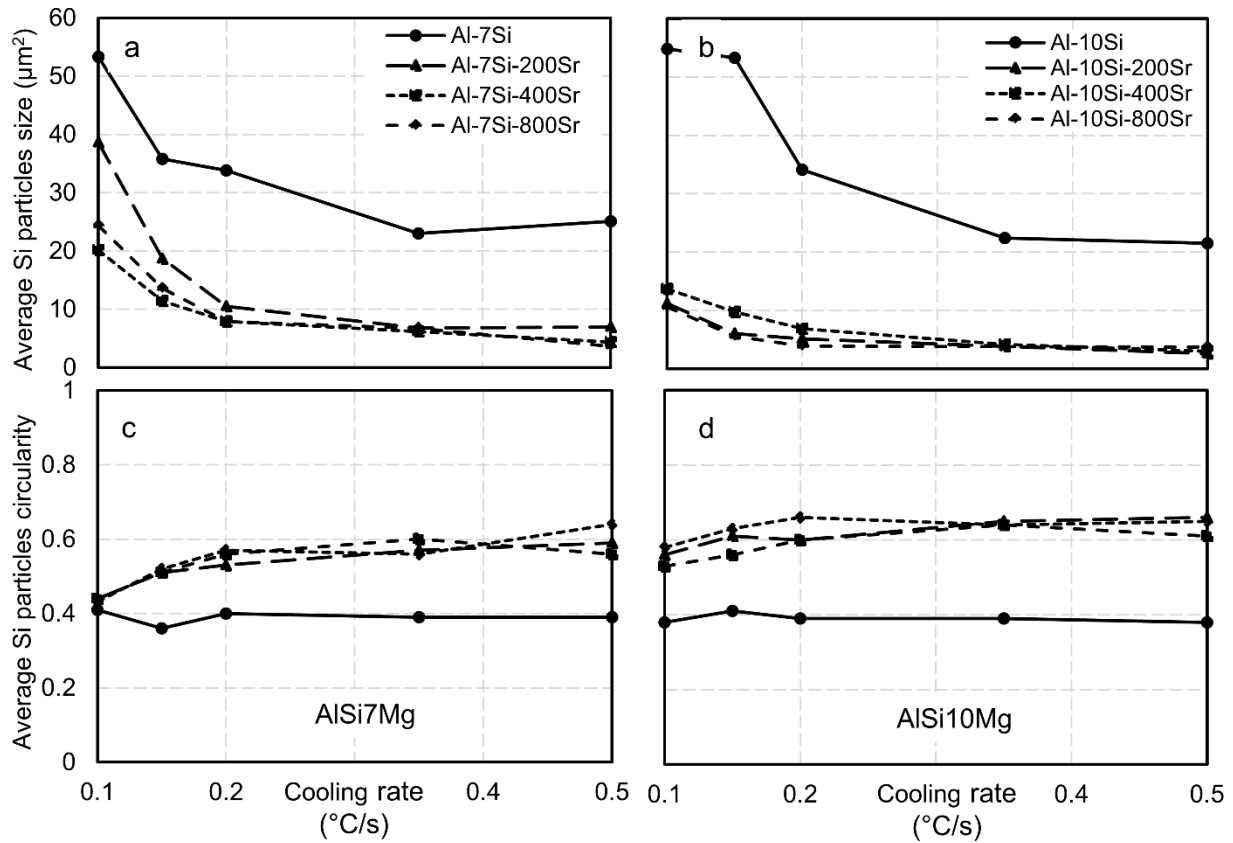


Figure 7-3- Influences of cooling rate and strontium addition on Si size for a) AlSi7Mg alloy and b) AlSi10Mg alloy, circularity of Si particles for c) AlSi7Mg alloy and d) AlSi10Mg alloy [45].

Previous studies on bulk samples suggested that strontium was not effective for modifying Al-Si alloys at low cooling rates. Ferdian et al. reported that cooling rates above 0.7 °C/s were necessary for effective modification of Al-7Si with Sr [105]. Wu et al. suggested an increase in Sr content to 700 ppm as a solution for modifying the alloy with the eutectic composition produced at low cooling rates [106]. However, the present study demonstrates that despite the cooling rates of less than 0.5 °C/s during solidification of open-cell foams, the eutectic silicon phase can be modified by strontium. For the Al-7Si alloy, the addition of 400 ppm Sr, and for the Al-10Si alloy, the addition of 200 ppm Sr, was sufficient to achieve the minimum silicon particle size. This suggests that modification at low cooling rates is more efficient when solidification occurs within confined volumes, such as the micrometer-sized struts of foams.

The main differences between foam and bulk castings can be grouped into two key aspects: 1- precautions in the casting process and 2- solidification condition. As discussed earlier, to completely fill the mold cavity, some precautions are required, such as keeping the mold at an elevated temperature, even higher than the liquidus

temperature of the alloy, or forcing the melt to flow within the fine strut structure by over- or under-pressure [44,130]. As shown in section 6.3, high mold temperature limits exogenous nucleation at the mold surface and single aluminum grains form inside the struts. Furthermore, solidification takes place in a very limited volume, which means that grain growth, phase formation, and segregation are confined to a limited space.

According to the studies on the modification mechanisms [103,164], the co-segregation of the modifier atoms (M) with Al and Si atoms leads to the formation of different types of nanometer sized Al-Si-M clusters. These clusters cause twinning in Si crystals and restrict their growth. In confined spaces such as foam struts, long-range atomic diffusion is not possible during solidification, increasing the likelihood that segregating atoms will remain in the last solidifying eutectic zones. This localized enrichment enhances the formation of Al-Si-Sr clusters and leads to a more efficient effect of Sr in modifying the Si phase in cast foams.

7.1.2. Mechanical properties and foam deformation behavior by modification

Figure 7-4 shows the compression stress-strain curves for one specimen per alloy. These curves show the foam deformation behavior under unmodified and modified conditions. The corresponding average values of the normalized peak stress, the normalized plateau stress, and the quasi-elastic gradient with change of Sr content in Al-7Si and Al-10Si foams are summarized in Figure 7-5.

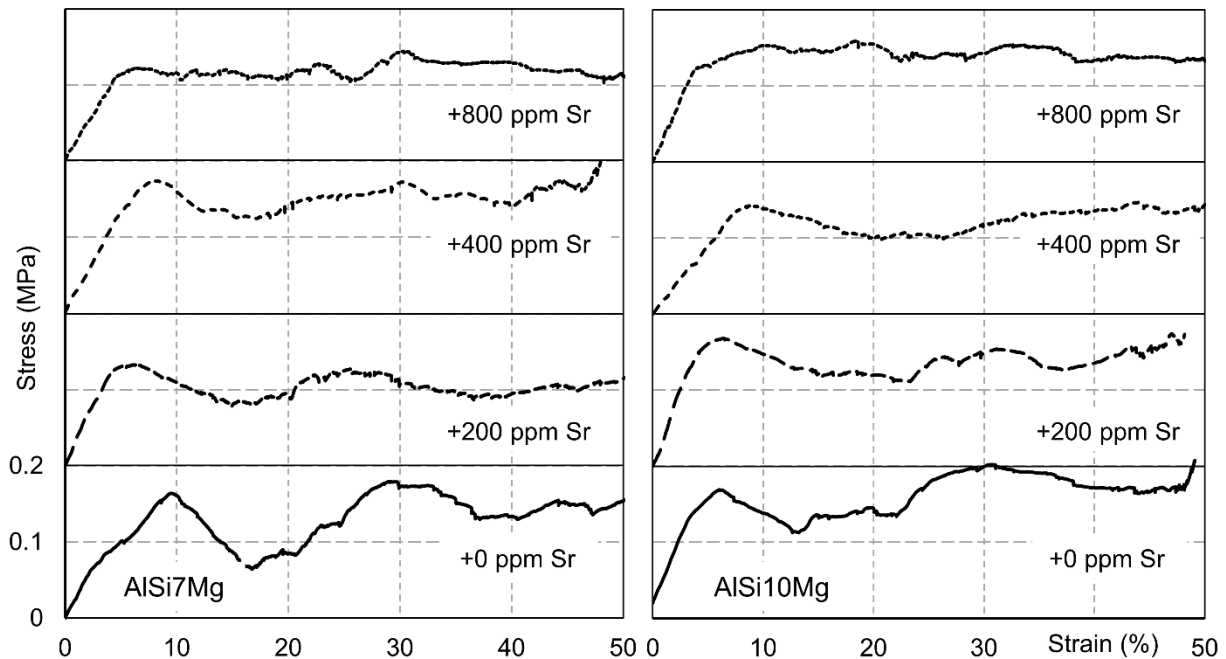


Figure 7-4- Influence of the Sr addition in different amount on the compressive stress-strain behavior of AlSi7Mg and AlSi10Mg Foams [45].

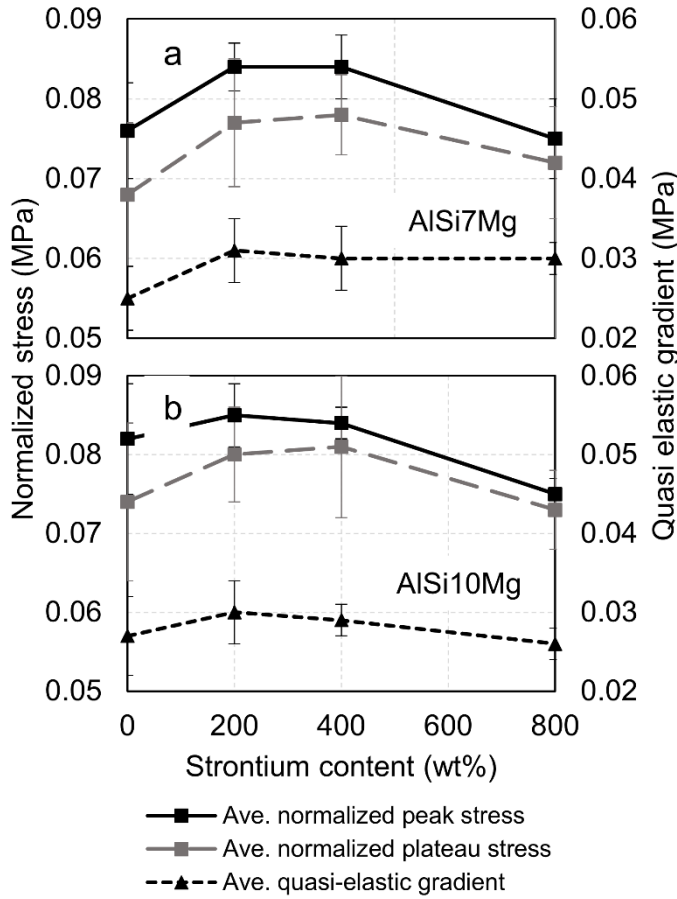


Figure 7-5- Average normalized peak stress, plateau stress, and quasi-elastic gradient for alloys a) AlSi7Mg and b) AlSi10Mg by change of strontium content [45].

For both Al-7Si and Al-10Si samples, a sudden and significant stress drop occurred after reaching the peak stress. In contrast, the Sr-modified samples exhibited much milder stress drops. With increasing Sr content, the stress drops were progressively eliminated, and the foams displayed more uniform deformation behavior with reduced fluctuations in the stress-strain curves.

To quantify the fluctuations within the plateau regime of each stress-strain curve, the total sum of squares (TSS) was calculated for the differences between the stress values and the plateau stress, starting from the peak stress up to the point where the strain reached 40%, as described in Equation 11 [165]. This statistical method is used to assess the dispersion of data points from the mean value, in this case, from the plateau stress.

$$TSS \text{ of stress} = \sum_{i=\text{peak stress}}^{\text{strain of 40\%}} (\sigma_i - \sigma_{pl})^2 \quad \text{Equation 11}$$

The average values of the TSS are reported for all compositions in Table 7-3. The alloys without strontium were found to have high TSS, which is a clear indication of large fluctuations in their stress-strain curves. In contrast, the stress-strain curves of alloys with 800 ppm strontium exhibited the lowest fluctuations. The lower

standard deviations of the TSS for the modified alloys indicate more homogeneous deformation, with the modified alloys exhibiting more consistent behavior compared to the unmodified ones.

Table 7-3- Average of the total sum of squares (TSS) for the differences between the real stress values and the plateau stress (standard deviations in parentheses) [45]

Alloy	Al-7Si				Al-10Si			
	0 Sr	200 Sr	400 Sr	800 Sr	0 Sr	200 Sr	400 Sr	800 Sr
Average TSS of stress	4.2 (2.2)	3.1 (1.4)	2.9 (1.4)	1.2 (0.7)	3.7 (4.2)	1.5 (1.2)	2.6 (1.8)	0.9 (0.4)

Stress fluctuations in the compression stress-strain curve is an indication of brittle fracture of the struts [10]. The accumulation of eutectic melt and consequently coarse silicon particles on the strut surface in the unmodified state (Figure 7-6a) causes brittle fractures during the compression test. The addition of Sr reduced stress oscillations and TSS values of the stress-strain curves, indicating a more ductile foam deformation and improved energy absorption efficiency. This is achieved by reducing the size of the silicon particles and improving their distribution. Unlike the unmodified condition, dendritic aluminum grains appear on the strut surfaces with Sr addition (Figure 7-6b), confirming a better silicon distribution. Consequently, the brittle surface layer becomes less dominant, allowing for more ductile strut deformation.

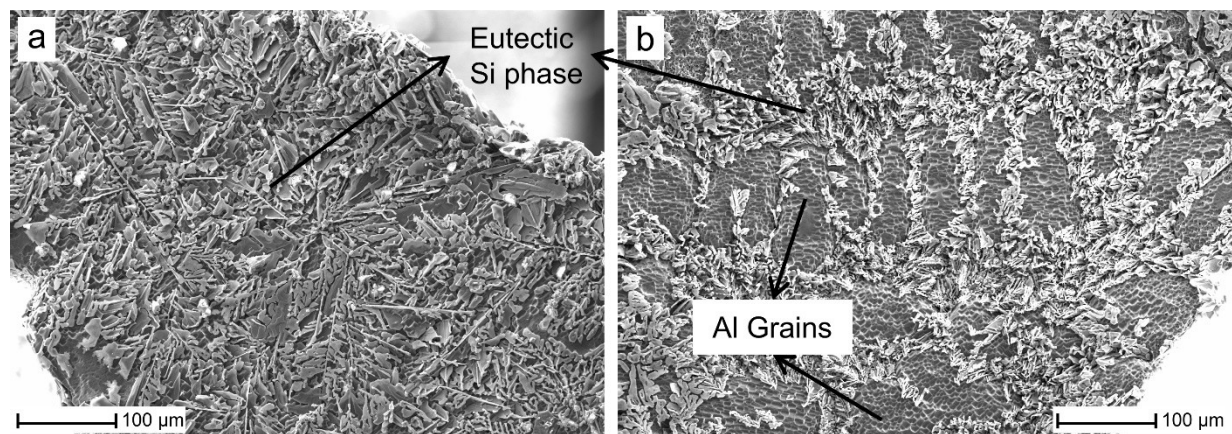


Figure 7-6- 3D structure of eutectic silicon phase on the strut surface, a) AlSi7Mg: coarse silicon particles cover the whole surface, b) AlSi7Mg400Sr: aluminum grains together with the fine silicon particles [45].

Although the addition of 800 ppm Sr resulted in the least fluctuation in the compression stress-strain curves of the foams, it negatively impacted the

mechanical properties. When considering all measured values at the foam level, the best results were achieved with the addition of 200 ppm and 400 ppm Sr in both alloys (Figure 7-5). However, an optimum Sr level between these two contents cannot yet be determined.

7.1.3. Strut-level properties by modification

The results of the micro-tensile tests on struts of the foams AlSi7Mg, Al7SiMgSr200, AlSi7MgSr400, and AlSi7MgSr800 are shown in Figure 7-7. Average values of UTS, uniform deformation strain, and fracture strain of struts together with their standard deviations are reported in Table 7-4.

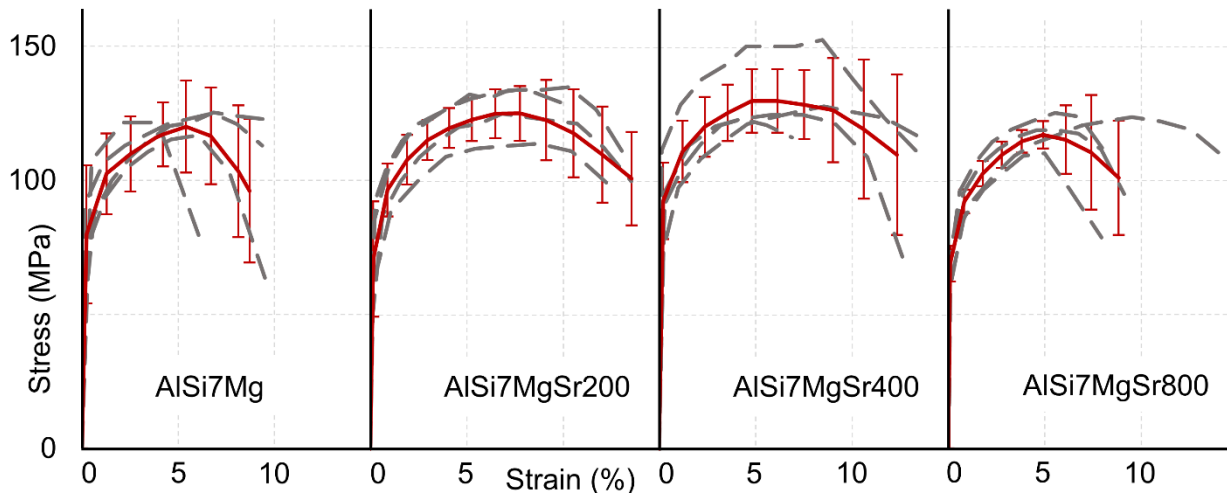


Figure 7-7- Stress-strain curves from micro-tensile tests of single struts by addition of strontium [45].

Table 7-4- Average values of UTS, uniform deformation strain, and fracture strain of single struts for AlSi7Mg alloy with different Sr content (standard deviations in parentheses) [45]

Alloy	AlSi7Mg			
	0 ppm	200 ppm	400 ppm	800 ppm
UTS (MPa)	124.8 (7)	127.2 (8.4)	132.3 (12)	119.6 (5.9)
Uniform deformation strain (%)	5.4 (1.1)	8.5 (1.2)	7.2 (1.4)	6.4 (1.4)
Fracture strain (%)	9.3 (3.5)	12.7 (1.6)	12.7 (1.5)	11.5 (2.7)

The positive influence of modification is also evident in the mechanical properties of the struts. Tensile strength and elongation at fracture improved with the addition of Sr up to a concentration of 400 ppm. However, the addition of 800 ppm Sr led to

a deterioration in these properties, resulting in a decrease in both tensile strength and elongation.

Previous studies on bulk castings indicated that the addition of Sr above a certain level leads to the formation of detrimental intermetallic compounds (like $\text{Al}_2\text{Si}_2\text{Sr}$) [166] and casting porosities (especially at low cooling rates) [167], which reduce the mechanical properties of the alloy. For instance, Zamani et al. [166] showed that the addition of 486 ppm Sr decreased the mechanical properties of AlSi9Cu3Fe alloy, highlighting the need to balance modification benefits with the risk of over-modification. Their findings support the idea that each alloy–casting condition combination has a unique Sr threshold, beyond which detrimental effects outweigh microstructural refinement.

In the present work on aluminum foam casting with AlSi7Mg and AlSi10Mg alloys, similar trends were observed. The addition of 800 ppm Sr appears to induce a decline in mechanical performance. A Sr content of 400 ppm was found to offer an optimal balance, providing a fine, fibrous Si morphology without triggering the formation of harmful intermetallic phases or promoting mechanical properties reduction.

7.2. Precipitation hardening

In contrast to Al-Si alloys, where microstructural refinement relies heavily on eutectic modification, Al-Cu alloys offer an alternative path for fine-tuning through heat treatment. As a classic heat-treatable alloy system, Al-Cu enables the in-situ formation of strengthening precipitates during post-solidification thermal processing. This approach is particularly valuable for aluminum foams, where controlled precipitation at the micro- and nanoscale can enhance mechanical performance. By adjusting both the alloy composition and heat treatment parameters, tailored mechanical properties can be achieved even in structures formed under low cooling rates. For this study foam samples type E are produced according to the chemical composition of Table 7-5 to study the microstructure evolution by aging and the corresponding mechanical properties.

Table 7-5- Chemical composition and casting condition of Al-Cu alloys

Melt ID	Si (wt%)	Fe (wt%)	Cu (wt%)	Melt temp. (°C)	Mold temp. (°C)
Al-4Cu	0.049	0.079	3.94	733	700
Al-5Cu	0.041	0.083	4.91	730	

7.2.1. Al-Cu foams in the as-cast condition

In the case of heat treatable aluminum alloys, a homogeneous dissolution of the precipitating atoms in the microstructure should be present prior to hardening in order to make optimum use of precipitation hardening. The available hardening mechanisms can be impaired by micro-segregation during solidification, which affects both the homogenization and the hardening behavior.

The severity of microsegregation in Al-Cu cast alloys, demonstrated by the phase fraction of non-equilibrium Al_2Cu in the as-cast condition, strongly depends on the cooling rate during solidification [168,169]. Figure 7-8a demonstrates the microstructural characteristics of the as-cast struts for Al-5Cu alloy. The non-equilibrium Al_2Cu (θ) phase grew into different morphologies of globular eutectic (GLE), elongated eutectic (ELE), globular blocky (GLB), and elongated blocky (ELB). Figure 7-8b shows the non-equilibrium θ area fractions in the as-cast alloys measured through metallography study.

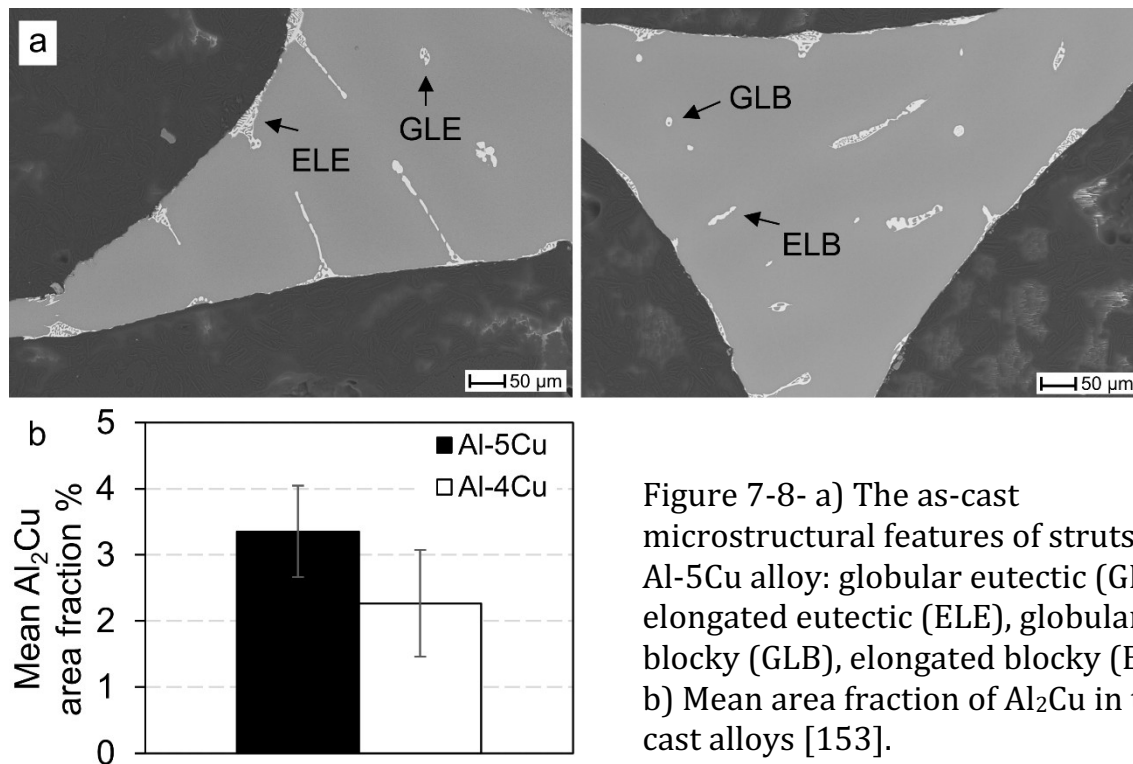


Figure 7-8- a) The as-cast microstructural features of struts for Al-5Cu alloy: globular eutectic (GLE), elongated eutectic (ELE), globular blocky (GLB), elongated blocky (ELB), b) Mean area fraction of Al_2Cu in the as-cast alloys [153].

High microsegregation levels can be attributed to the solute partitioning at low cooling and solidification rates. However, dendrite arm coarsening, which occurs during slow cooling after solidification, can partially homogenize the segregation profile within the solid phase [168,170]. During coarsening, fine dendrite arms that formed early in solidification and contain lower solute content gradually dissolve, while coarser arms thicken. This process redistributes solute atoms within the solid, reducing local compositional variations. Although this can moderate residual

segregation, it does not significantly eliminate the non-equilibrium θ phase formed during solidification. Thermodynamic simulation provides the chance to understand the copper microsegregation in our Al-Cu alloys after solidification.

Figure 7-9 illustrates the simulated vs. measured Cu microsegregation profile of Al-4Cu alloy, as well as the region where the elemental line scanning was performed. The brighter areas in the SEM image correspond to the θ phase, while the darker areas correspond to the α -Al matrix. Al_2Cu formed in the region between the dendrite arms, therefore the distance between the bright areas was considered as the SDAS, which was equal to $116\text{ }\mu\text{m}$. The simulated concentration profile agreed well with the experimental results in both qualitative and quantitative manners.

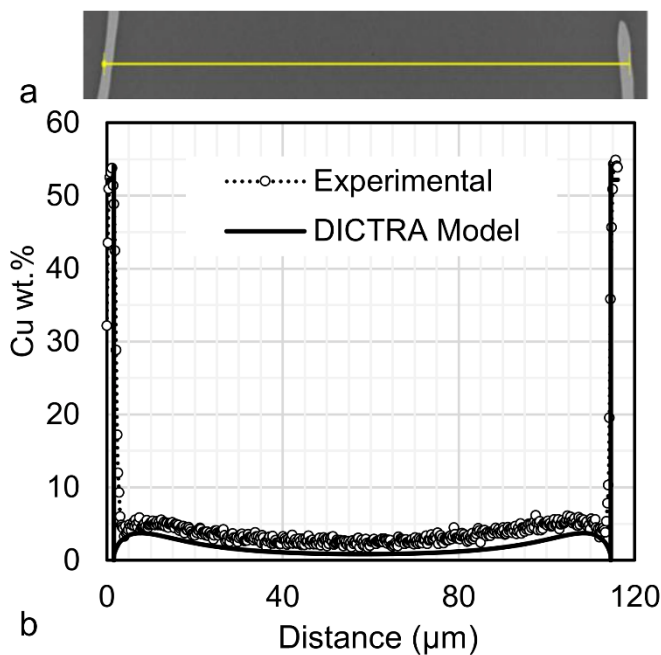


Figure 7-9- Cu microsegregation profile inside a secondary dendrite arm of Al-4Cu: a) the elemental line scanning region within one dendrite arm, b) the measured vs. predicted Cu concentration profiles [154].

After solidification, the cast foams cool slowly to room temperature, which influences microsegregation behavior, as can be realized qualitatively from the Cu microsegregation profile. The shape of the segregation curve suggests atomic flux in the reverse direction, indicating back diffusion [169]. This process contributes to partial microstructural homogenization. However, the slow cooling also promotes growth in dendrite arm spacing, resulting in a coarser microstructure that can significantly counteract any homogenization achieved through back diffusion. Furthermore, since Cu diffuses sluggishly within the Al matrix, the extent of homogenization due to back diffusion is likely to remain limited.

7.2.2. Homogenization and solution annealing

In order to dissolve the observed non-equilibrium phase prior to precipitation hardening treatment, a solution annealing is necessary to homogenize the Cu concentration throughout the strut. To determine the optimal temperature for this

treatment, thermal analysis of the melt was performed, as shown in Figure 7-10. In the first derivative of the cooling curve, the initial inflection point corresponds to the liquidus temperature and the nucleation of primary α -Al dendrites. The second inflection marks the nucleation of Fe-rich intermetallic phases, while the third point indicates the non-equilibrium eutectic transformation of the Al_2Cu phase, which occurs during the final stages of solidification.

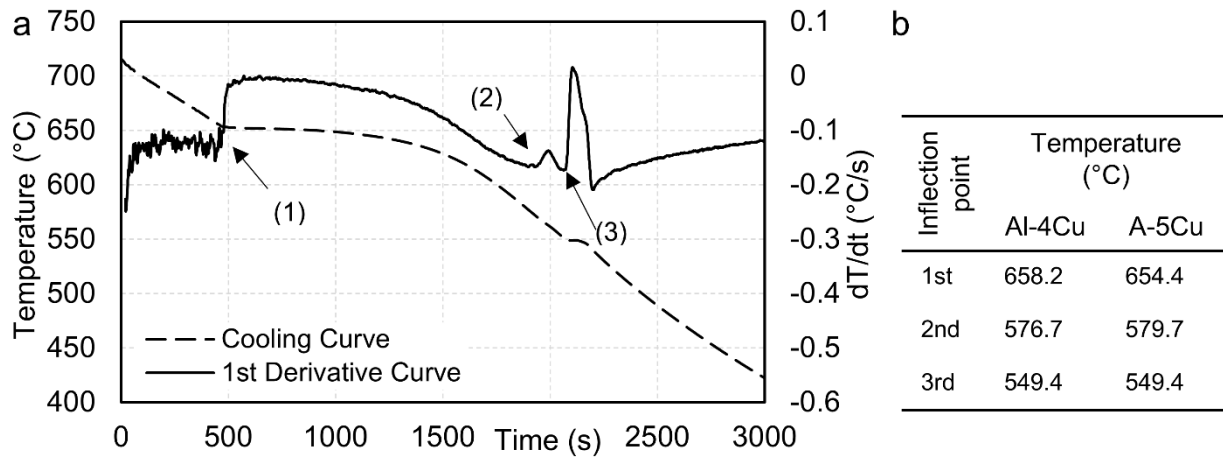


Figure 7-10- a) Thermal analyses from cooling curve of Al-5Cu alloy, b) the corresponding temperatures to the inflection points of Al-4Cu and Al-5Cu alloys, 1st point as liquidus temperature, 2nd point as nucleation of intermetallic phases, and 3rd point as non-equilibrium eutectic temperature [154].

Based on the thermal analysis, the solution annealing temperature must remain below the non-equilibrium eutectic temperature of 549.4 °C to prevent re-melting of the last solidified regions. Therefore, the foams were homogenized at 525 °C for different durations of 8, 16, and 24 hours. With increasing annealing time, the non-equilibrium θ -phase gradually dissolved, leaving only the iron-rich intermetallic phase. According to the thermal analysis, the formation temperature of the iron phase is above 570 °C; hence, its dissolution cannot be achieved through solution annealing within the safe temperature range.

7.2.3. Aging and precipitation hardening

Based on previous studies on the aging behavior of bulk Al-Cu binary alloys [171,172], aging temperatures of 160 °C, 200 °C, and 300 °C for 24 hours were selected for the cast foams. The resulting microstructures after solution annealing at 525 °C for 24 hours followed by aging at these temperatures are shown in Figure 7-11. No detectable precipitates were observed by SEM in the samples aged at 160 °C. Aging at 200 °C resulted in the formation of fine precipitates in the nanometer range, likely corresponding to the metastable θ' phase. In contrast, aging at 300 °C produced coarser precipitates exceeding 2 μm in size, representative of the equilibrium θ phase.

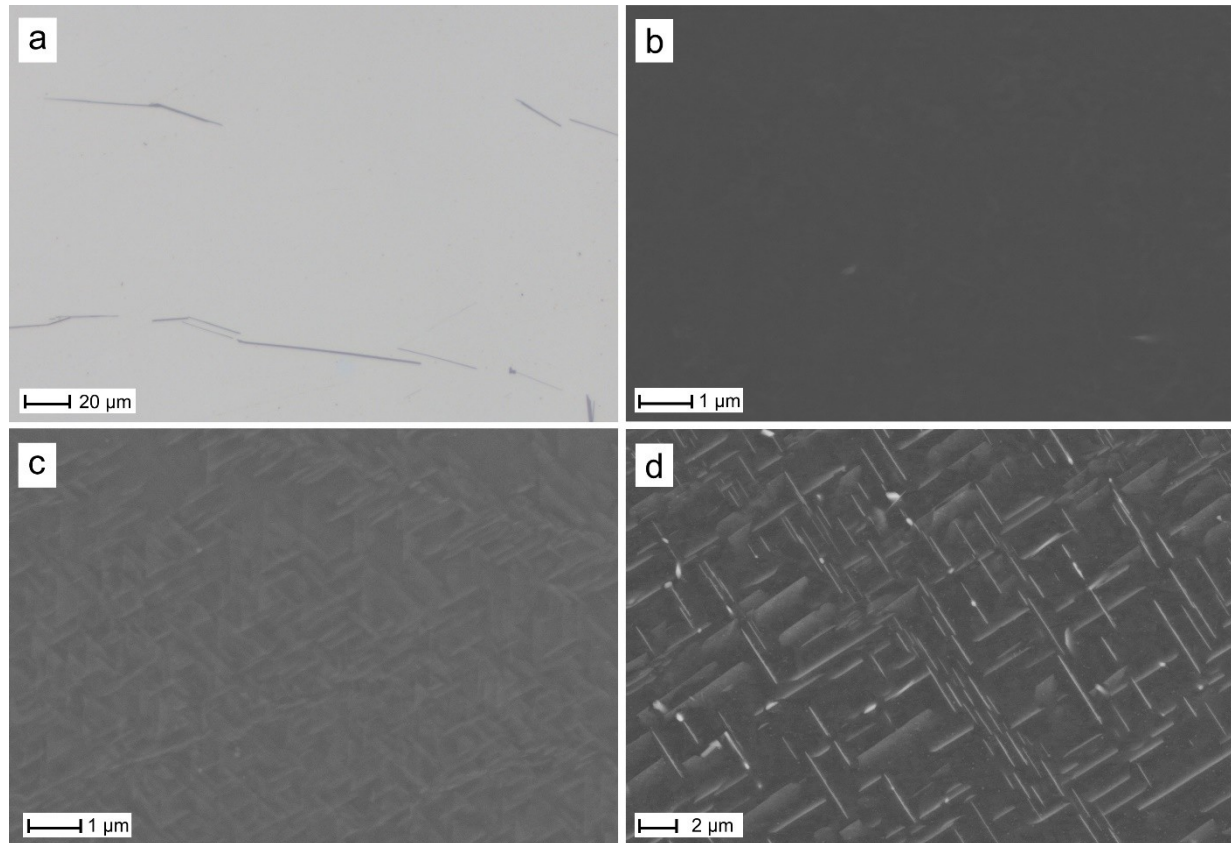
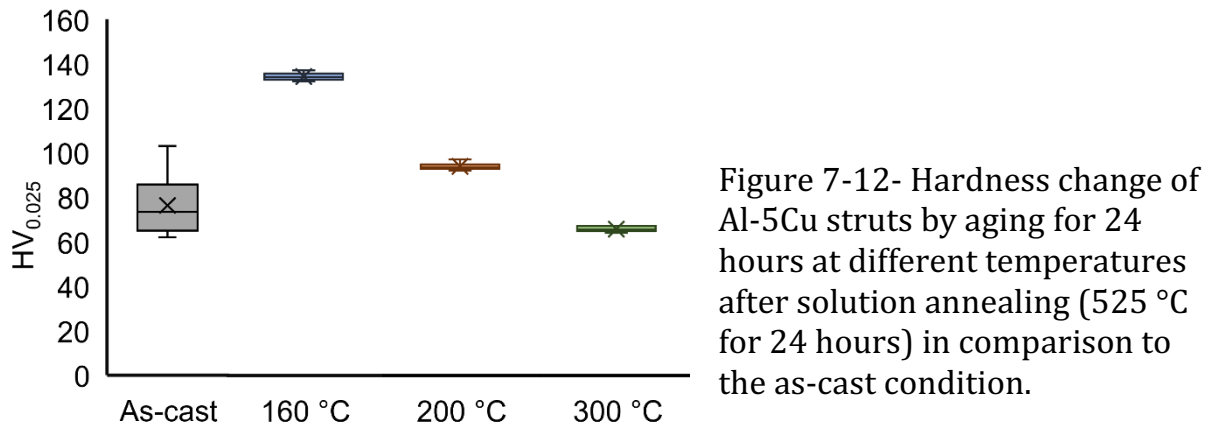


Figure 7-11- Micrographs of Al-5Cu alloy at different conditions: a) after solution annealing at 525 °C for 24 hours (optical microscope), b) aged at 160 °C for 24 hours (SEM), c) aged at 200 °C for 24 hours (SEM), and d) aged at 300 °C for 24 hours (SEM).

Microhardness measurements on the vertically oriented struts of the Al-5Cu alloy revealed clear changes as a function of aging temperature, as shown in Figure 7-12. In the as-cast condition, the matrix exhibited a hardness of 73 ± 17.3 HV, with significant scatter due to microstructural inhomogeneity. Aging at 160 °C produced the highest hardness, reaching 134 ± 2.3 HV. However, increasing the aging temperature to 200 °C and 300 °C led to reduced hardness values of 94 ± 2.7 HV and 68 ± 1.6 HV, respectively. These variations in hardness align well with the microstructural observations. Aging at 160 °C promotes the formation of very fine, nanoscale precipitates, likely GP zones, which enhance hardness but remain undetectable in SEM imaging. Increase of aging temperature results in increase of precipitate size and consequently reduces the hardness.



7.2.4. Mechanical properties by heat treatment

Figure 7-13 presents the normalized compressive stress–strain curves of Al-5Cu foam samples type E (as described in Table 5-2) in both the as-cast and heat-treated states (three samples per condition). Compared to Al-Si alloy foams, the Al-Cu foams exhibit reduced serration and a noticeable degree of strain hardening. The suppression of serrations can be attributed to two factors: first, the relatively large sample size ($50 \times 50 \times 50 \text{ mm}^3$), which helps to average out local fluctuations in deformation; and second, the absence of brittle particles, such as eutectic Si, in the microstructure. Without such stress-concentrating features, the deformation process proceeds more homogeneously, contributing to a smoother transition from quasi-elastic to plastic deformation and a more gradual shift from the plateau region to densification.

The average normalized stresses at the transition and plateau regions (see Table 7-6) highlight the influence of heat treatment on foam performance. The highest mechanical properties were achieved in the samples aged at 160 °C that can be attributed to the precipitation of fine, uniformly distributed precipitates. However, aging at 300 °C results in coarsening or over-aging of the precipitates, leading to a reduction in strengthening effectiveness. As a result, the mechanical performance of the 300 °C-aged samples closely resembles that of the as-cast condition. This similarity is probably due to the fact that both conditions exhibit relatively coarse and sparsely distributed intermetallic phases that offer only limited resistance to plastic deformation. In the as-cast condition, the hardening potential is inherently low due to the lack of refined precipitates, while in the over-aged condition, the precipitates are too large and too far apart to contribute effectively to strength.

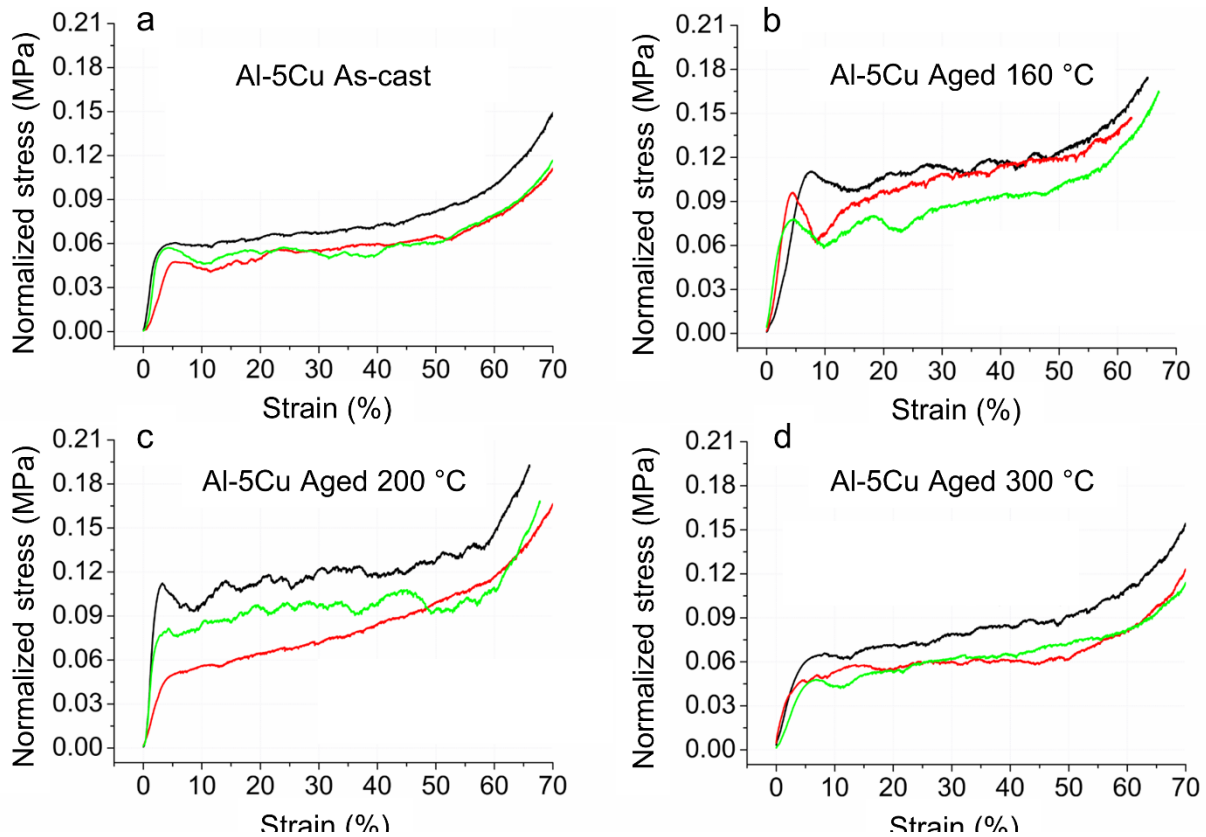


Figure 7-13- Normalized compressive stress-strain curves of Al-5Cu foam samples type E for as-cast and aged conditions.

Table 7-6- Average normalized transition stress (elastic to plastic) and plateau stress for Al-5Cu foam samples for as-cast and aged conditions

	As-cast	Aged at 160 °C	Aged at 200 °C	Aged at 300 °C
Ave. normalized transition stress	0.05	0.09	0.08	0.05
Ave. normalized plateau stress	0.10	0.17	0.16	0.11

7.3. Role of the secondary phases

In the case of homogeneous microstructure free from brittle particles, as in the case of heat-treated Al-Cu alloys, the mechanical properties of the foam correlate well with the fundamental material characteristics, such as strut hardness (Figure 7-14). This relationship follows the same principle observed in bulk materials, where microstructural refinement through heat treatment improves mechanical performance in a predictable way. The absence of embrittling phases enables stable and controlled deformation, allowing the mechanical response of the foam to be effectively optimized through microstructural tuning.

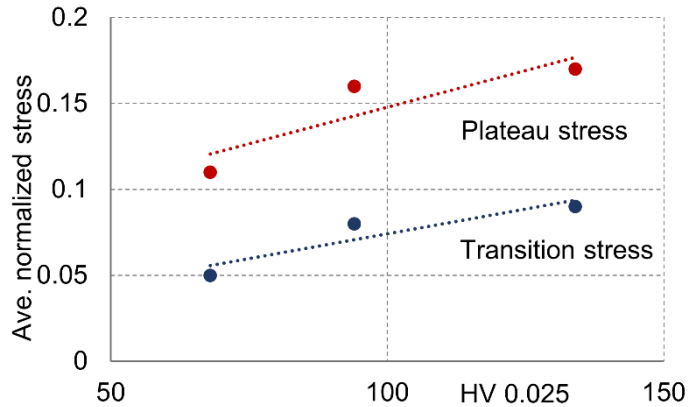


Figure 7-14- Direct influence of hardness of Al-5Cu struts on mechanical properties of foams: normalized plateau and transition stresses.

In contrast, the presence of brittle Si particles in Al-Si foams leads to complexity and variability in their deformation behavior. The influence of Si particles depends on several factors, including their size, location, and distribution within the struts. When the Si particles are fine and uniformly dispersed, their detrimental effect can be partially moderated. However, larger or clustered particles serve as stress concentrators, triggering early crack initiation and abrupt failure. This results in less predictable mechanical behavior at the foam level, making performance optimization more challenging compared to Al-Cu systems.

A decisive difference arises when considering microstructural uniformity versus heterogeneity. In homogeneous materials, mechanical properties can be reliably tailored through microstructural control, similar to what is achievable in conventional bulk castings. However, when a secondary phase (e.g., Si particles) is present and its size approaches or exceeds critical structural dimensions such as the strut thickness, the material can no longer be treated as homogeneous. In these cases, the overall mechanical behavior is influenced not only by the base alloy matrix but also by the properties of the secondary phase. Variations in particle size, morphology, and spatial distribution introduce substantial inconsistency in mechanical response, making it difficult to establish a direct correlation between foam-level behavior and fundamental material properties.

This heterogeneity becomes particularly critical in open-cell foams, where individual struts serve as the primary load-bearing elements. Unlike solid materials, where stress is distributed more uniformly across the entire volume, foams are highly sensitive to localized weak points within the struts, which can directly and immediately impact mechanical performance. In Al-Si foams, for instance, brittle struts containing concentrated Si-rich phases are prone to abrupt failure, often visible as sharp stress drops in the compressive stress-strain curve. In contrast, Al-Cu foams with a refined and homogeneous microstructure exhibit stable, progressive deformation, minimizing sudden fractures and stress instabilities.

Ultimately, the ability to precisely tailor the mechanical properties of open-cell cast foams relies on effective microstructural control. In materials with a uniform microstructure, properties can be systematically adjusted through processing steps such as solution treatment and aging, resulting in a predictable and reproducible behavior. In contrast, when coarse or unevenly distributed secondary phases are present, mechanical performance becomes naturally variable. This variability necessitates additional considerations in both alloy design and foam processing to reduce unpredictability and achieve optimal mechanical behavior.

7.4. Conclusion: Tuning the second phase microstructure

When hard and coarse silicon particles are present in the microstructure of binary aluminum–silicon alloys (e.g., Al-7Si and Al-10Si), modification can improve the mechanical properties to some extent:

- Strontium is an effective modifier of silicon in cast aluminum foams, but only when the cooling rate exceeds 0.3 °C/s.
- To meet this cooling condition, the foam should be placed in a favorable location within the casting flask (ideally close to the flask wall).
- In confined solidification spaces, such as the micrometer-sized struts of foams, the modification appears more efficient at lower cooling rates compared to bulk samples.
- Sr addition reduces fluctuations in the plateau region of the stress–strain curves and promotes ductile deformation. The addition of up to 400 ppm Sr improves the compressive properties of the foam and the tensile properties of the struts. However, higher Sr contents lead to over-modification and a decline in mechanical performance.

For fine, in-situ precipitates formed through heat treatment, more ductile deformation is observed, along with reduced serration and stress drops in the plateau region of the stress–strain curves. In the case of binary Al-Cu alloys (e.g., Al-5Cu):

- Homogenization at 525 °C effectively dissolves non-equilibrium Al₂Cu phases formed during solidification; however, iron-rich intermetallics remain undissolved.
- Aging at 160 °C leads to the formation of nanoscale precipitates, increasing strut hardness as well as the transition and plateau stresses at the foam level.
- The mechanical properties of aluminum alloys foam with uniform microstructure and ductile deformation can be reliably tailored through microstructural control, similar to conventional bulk castings.

8. Proof of Concept: Hybrid Investment Casting of Lattice Structure

8.1. Cast lattice structures

In the first step in order to establish the process chain, 3D-printed patterns from vat polymerization were integrated into the block mold casting process to produce aluminum alloy lattice structures with the alloy AlSi10MgMn. For this purpose, BCC_0.7, Diamond, and Kelvin structures were produced according to Table 8-1.

Table 8-1- Chemical composition and casting condition for the production of cast lattice structures out of Al10SiMgMn alloy in the initial study phase (3 specimens per condition)

Sample ID	Resin of pattern	Strut diameter (mm)	Si wt%	Fe wt%	Mg wt%	Mn wt%	Melt temp. (°C)	Mold temp. (°C)
BCC_0.7E	Epic	0.73	9.93	0.09	0.30	0.28	683	700
BCC_0.7P								
Diamond	PIC 100	0.74	9.89	0.08	0.30	0.29	681	700
Kelvin		0.68	9.91	0.08	0.30	0.28	686	700

8.1.1. BCC structure

The BCC structure is widely recognized in the design of lattice structures as it offers an excellent balance between high stiffness and low weight, making it ideal for applications focused on weight reduction [59,60]. Its relatively simple geometry enables not only efficient production and excellent printability, but also easy integration into complex systems. It was therefore chosen initially to investigate the differences in the behavior of the various resin types in this study.

The BCC_0.7 lattice structures were produced with printed patterns from Epic green and PIC 100 resins. The results of the compression test of the BCC samples with both casting resins (Figure 8-1) showed that they provided the same quality of lattice structures with similar mechanical properties and deformation behavior. The firing process up to 750 °C to burn out the patterns and fire the molds was efficient to create a clean and suitable mold cavity for casting. Residues in the mold cavity could otherwise lead to variations in lattice deformation. The homogeneous and consistent deformation behavior observed therefore indicates both the successful removal of resin components during firing and the equivalent performance of the two resins. From this point onward, patterns were produced only with PIC100 resin.

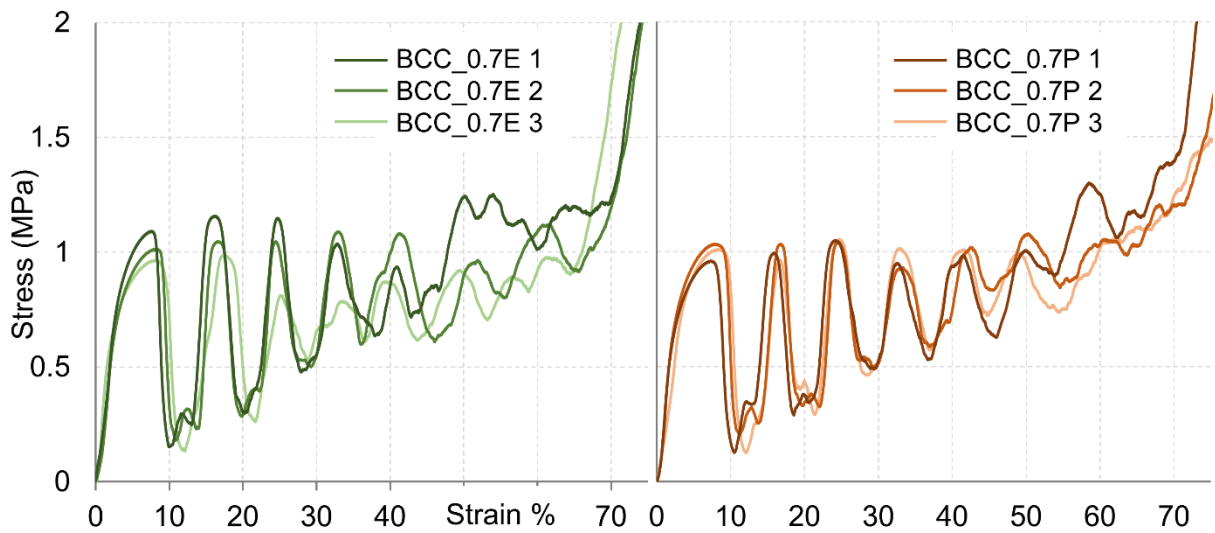


Figure 8-1- Stress-strain curves for BCC lattice structure: printed patterns of Epic green and PIC 100 resins [173].

During compression, there were several stress buildups and releases, which led to fluctuations in the stress-strain curves. Figure 8-2 shows the average stress-strain curve for all six BCC_0.7 specimens, along with a standard deviation band. The deformation states of a representative specimen are shown at three peak and three valley stress points to illustrate the course of structural collapse during loading.

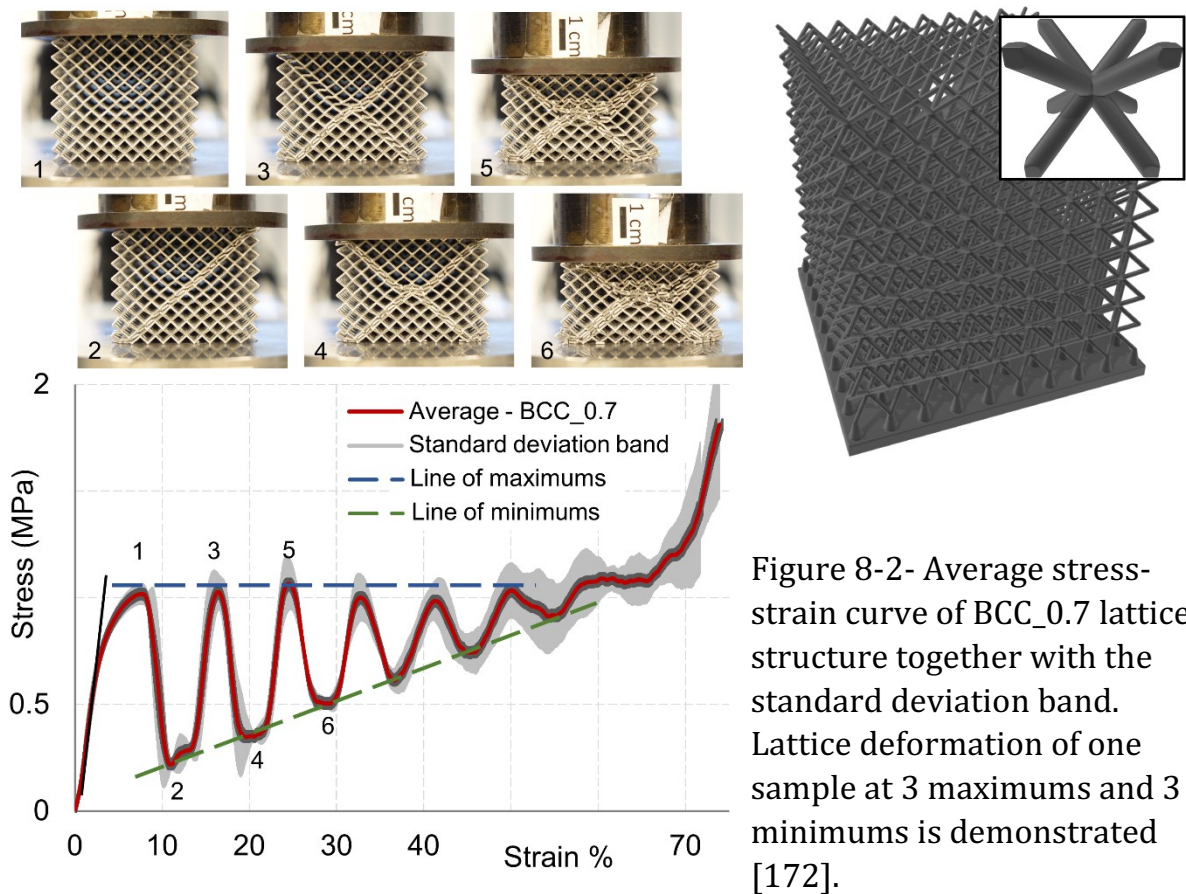


Figure 8-2- Average stress-strain curve of BCC_0.7 lattice structure together with the standard deviation band. Lattice deformation of one sample at 3 maximums and 3 minimums is demonstrated [172].

Closer examination of the BCC samples during compression revealed the formation of shear bands at a 45° angle, which is a characteristic feature of the mechanical response of this lattice structure. The initial shear band developed across the entire diagonal of the cube. Since there were no additional nodes in the longitudinal direction to interrupt the load path, the shear band extended through the full length of the structure. Following the quasi-elastic region, a slight plastic deformation occurred, indicated by the deviation from the linear elastic slope (represented by the black solid line). However, due to the limited ductility of the material, the row of struts along the shear band fractured simultaneously, leading to the first noticeable stress drop.

After the first stress drop, subsequent stress build-ups were observed. In these cases, the stress dropped again after reaching a maximum, but without any indication of plastic deformation. The primary difference between the first stress peak and the following ones lies in the amount of material engaged in load-bearing. At the initial stress maximum, all rows of struts were intact and contributed to supporting the load, allowing for a greater potential for plastic deformation. However, as struts progressively collapsed after each stress drop, less material remained to carry the load, thereby reducing the structure's capacity for further plastic deformation.

The maximum stresses (indicated by the blue dashed line) remained relatively constant throughout the deformation process up to the onset of densification. This represents the maximum tolerable stress of the BCC structure; each time this value was reached, additional struts fractured. As the number of broken struts increased, the available free space for further compression diminished. Consequently, the stress build-up began earlier in each cycle, leading to a progressive increase in the minimum stress values (represented by the green dashed line).

8.1.2. Diamond structure

The Diamond structure is derived from the FCC lattice, with additional nodes positioned at one-quarter of the diagonal length from each corner. This configuration results in each node being connected by four struts, resembling the atomic arrangement found in diamond crystals. Due to a spatial shift in the node placement, the Diamond lattice features shorter struts and a denser node distribution compared to the BCC structure. Even with identical relative density and strut diameter, the shorter and more numerous struts of the Diamond lattice provide a more compact and coherent framework. The average stress-strain response of the Diamond structure, along with representative deformation images of a sample, is shown in Figure 8-3.

The formation of a shear band at approximately 45° was observed during the initial stage of plastic deformation in the Diamond specimens (highlighted in the red circle). However, due to the presence of four internal nodes within each unit cell, the Diamond structure restricts the propagation of shear bands across the entire length of the specimen. Instead, localized shear banding occurred, and the failure of the first set of struts led to a noticeable sliding between the two resulting segments of the specimen. As deformation progressed, these segments developed into two separate, misaligned sections. The stress exceeded 2 MPa before the struts fractured at the shear band location, followed by a sharp drop to a minimum stress level. The complexity of the unit cell and the spatial displacement of its nodes hindered the possibility of uniform deformation across the structure.

Unlike the BCC structure, the Diamond lattice did not exhibit noticeable fluctuations in the stress-strain curve. Instead, a continuous increase in stress was observed following the initial stress minimum, with no distinct plateau region. This behavior can be attributed to the short struts and the irregular collapse of cells across the sample, which allowed progressive compression of the struts. The resulting strain hardening effect is driven by the geometric interaction of structural elements rather than by the intrinsic properties of the material itself [47].

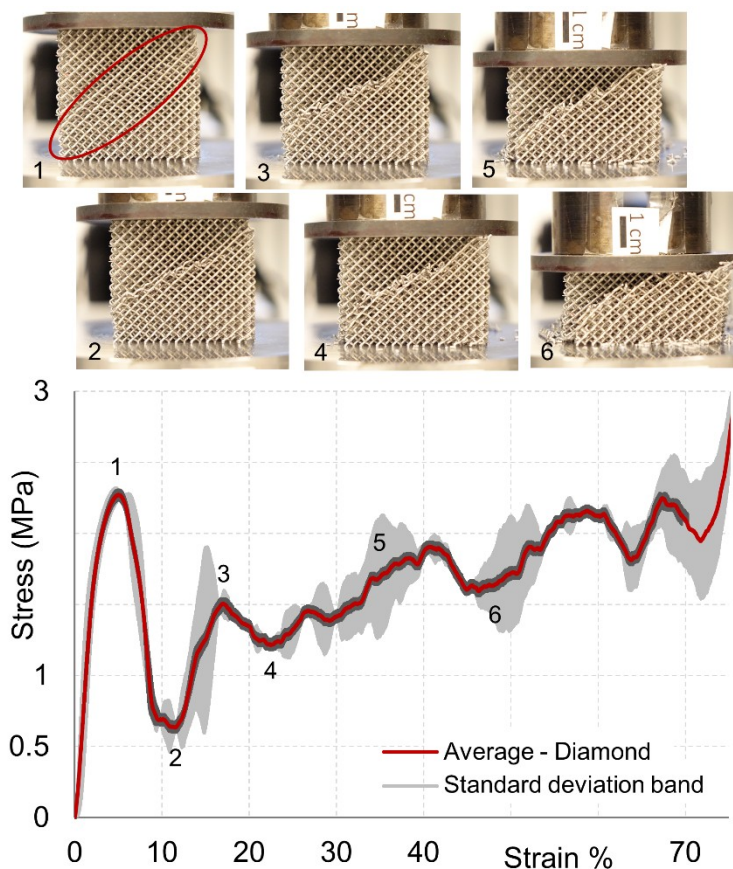


Figure 8-3- Average stress-strain curve of Diamond lattice structure together with the standard deviation band. Lattice deformation of one sample at 6 points is demonstrated [172].

8.1.3. Kelvin structure

The Kelvin structure is based on the tetrakaidekahedron, a 14-faced polyhedron that can fill space without gaps. In this unit cell design, some of the struts lie perpendicular to the compression loading direction within the base planes of the structure, while others are oriented at approximately 45° to the loading axis. This unique configuration results in a complex load distribution during deformation. The average stress–strain curve for the Kelvin lattice structure, along with representative deformation images of a Kelvin sample, is presented in Figure 8-4.

The deformation behavior of the Kelvin structure differed notably from that of the BCC and Diamond structures, beginning at an early stage (around 3% strain) with the fracture of struts in the top layer. As a result, unlike the other two structures, no distinct peak or minimum stresses were observed in the stress–strain curve. As deformation progressed, shear bands formed in localized regions (highlighted in red circles), and individual strut failures led to minor stress drops. However, the overall deformation was more uniform, and the stress–strain response exhibited a relatively flat plateau.

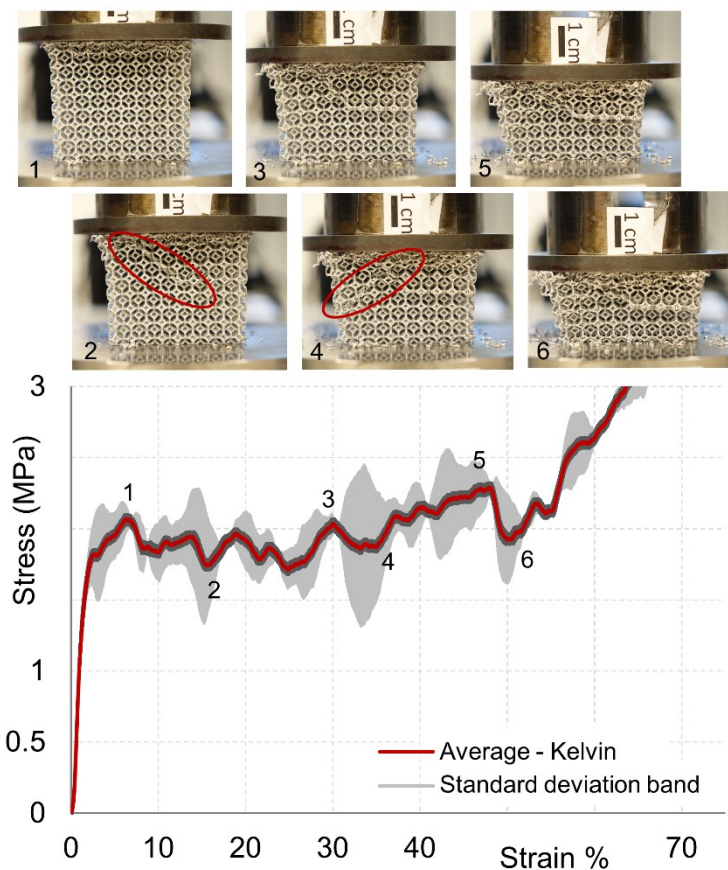


Figure 8-4- Average stress-strain curve of Kelvin lattice structure together with the standard deviation band. Lattice deformation of one sample at 6 points is demonstrated [172].

All mechanical properties of the three structures are gathered in Table 8-2. Due to the absence of pronounced fluctuations in the Kelvin structure, the plateau stress,

defined as the average stress after the elastic region, was higher than that of the other structures. Moreover, both the absorbed energy and, more importantly, the energy absorption efficiency of the Kelvin structure were relatively high. An energy absorption efficiency closer to 1 indicates a more effective energy-absorbing lattice structure. However, in the Kelvin structure, many struts are oriented more parallel to the compression direction and intersect sooner, leading to earlier densification. In contrast, the BCC and Diamond structures feature more angled struts, which delay densification and allow deformation to proceed to higher strains.

Table 8-2- Mechanical properties of lattice structure with BCC, Diamond, and Kelvin unit cells with the same relative density of around 10 %

Parameter from compression test	BCC	Diamond	Kelvin
Plateau stress (MPa)	0.71 (0.02)	1.47 (0.08)	1.98 (0.06)
Densification strain (%)	67.4 (2)	71.2 (0.7)	63.3 (0.2)
Energy absorption per volume till densification (MJ/m ³)	0.53 (0.04)	1.18 (0.01)	1.30 (0.04)
Energy absorption efficiency	0.74 (0.02)	0.73 (0.02)	0.82 (0.06)
Peak (deviation from elastic) stress (MPa)	1.02 (0.05)	2.27 (0.03)	1.89 (0.22)
Strain at the peak stress (%)	7.93 (0.28)	5.18 (0.35)	3.02 (0.69)
2nd Maximum stress (MPa)	1.04 (0.07)	-	-
3rd Maximum stress (MPa)	1.07 (0.05)	-	-
4th Maximum stress (MPa)	1.01 (0.06)	-	-
5th Maximum stress (MPa)	1.01 (0.05)	-	-
1st Minimum stress (MPa)	0.16 (0.03)	0.23 (0.01)	-
2nd Minimum stress (MPa)	0.30 (0.02)	-	-
3rd Minimum stress (MPa)	0.48 (0.02)	-	-
4th Minimum stress (MPa)	0.58 (0.04)	-	-
5th Minimum stress (MPa)	0.70 (0.08)	-	-

The peak and plateau stresses of the Diamond structure were more than twice as high as those of the BCC structure, highlighting the superior load-bearing capacity of the Diamond lattice. This also suggests that the unpredictable and random deformation behavior of the Diamond structure may contribute to its higher strength. On the other hand, the BCC structure exhibited the highest strain at peak

stress and therefore the longest quasi-elastic range. This indicates that longer struts and a more open structure promote greater elastic deformation.

8.1.4. Microstructure

The general features of the microstructure were consistent across all three structures. Micrographs of cast lattice structures, shown in Figure 8-5, reveal primary aluminum grains in a dendritic form, which appear as a white matrix. The eutectic silicon phase is present as dark gray particles between the dendrite arms and on the strut surfaces. Due to the presence of manganese as an alloying element, iron-containing phases were observed in the Chinese script morphology, appearing as a light gray phase, with no needle-like iron phase (beta phase) detected. Shrinkage porosity was more likely to be observed in the nodes and strut connections, as indicated by the red arrows. Microstructure parameters, such as the average silicon particle size and secondary dendrite arm spacing, are provided in Table 8-3.

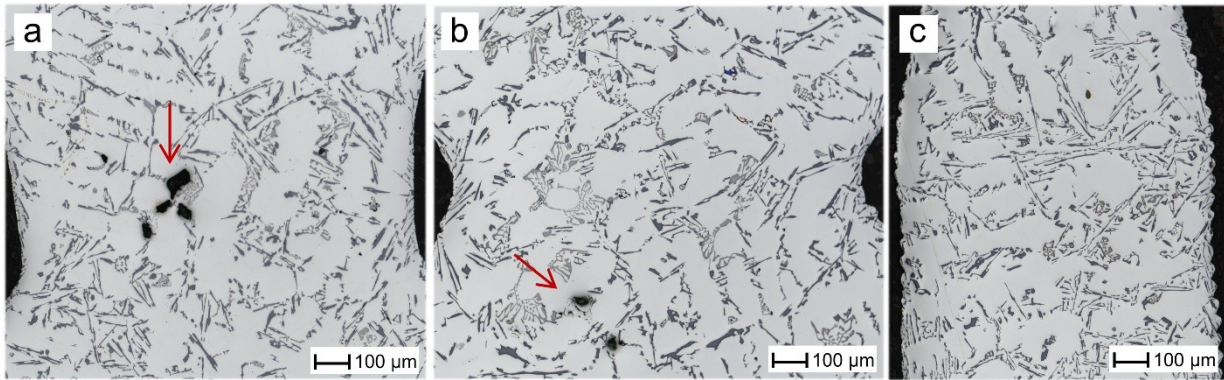


Figure 8-5- Microstructure of the cast lattice structures: a) BCC lattice at a nod, b) Diamond lattice at a nod, and c) Kelvin structure at a vertical strut [173].

Table 8-3- Average values of silicon particle size and secondary dendrite arm spacing for different lattice structures (standard deviations in parentheses)

Microstructure features	BCC	Diamond	Kelvin
Silicon particle size (μm^2)	277.5 (42.3)	293.4 (39.5)	237.4 (25.8)
Secondary dendrite arm spacing (μm)	87.4 (5.6)	89.6 (6.8)	83.5 (4.2)

8.2. Deformation behavior of cast lattice structure

The brittle deformation observed in the cast lattices can be attributed to the microstructural characteristics of the 10 wt% silicon alloy. The slow cooling rate in the block mold casting process led to the formation of large eutectic silicon particles, which reduced the ductility of the material [44,79]. In addition, the presence of

shrinkage porosities further hindered ductile deformation. Fracture typically occurred at the nodes (intersections of the struts), coinciding with areas of shrinkage porosity, highlighting their critical role in the failure mechanism.

The formation of shear bands and the localization of deformation at approximately 45° to the compression direction have been well-documented in previous studies [174,175]. This behavior is due to the alignment of the maximum shear stress planes with this orientation. Under compressive loading, peak shear stresses develop along the 45° inclined planes, promoting strain localization and facilitating the formation of shear bands. This phenomenon is further influenced by the bending and eventual failure of struts within the lattice.

As explained before and according to the Maxwell's criterion [16,22], if there are too few struts (negative Maxwell number) to balance the external forces without equilibrating bending moments at the nodes, bending stresses will form in the struts according to the bending-dominated behavior. Conversely, when there is a high number of struts surrounding the nodes (positive Maxwell number), external loads can be balanced by axial tension or compression in the struts, causing the structure to behave in a stretch-dominated manner [16,60]. All three lattice structures in this study are bending-dominated, meaning the deformation is driven by bending stresses rather than by axial loading. Simulation studies [142,176] have shown that the highest bending stresses occur at the midpoints of inclined struts, which should result in S-shaped deformation [18]. However, in the present work, fractures consistently occurred at the nodes rather than at the expected bending points. This behavior can be explained by the fact that, in cast lattices, the junctions represent the weakest regions due to material accumulation and hot-spot formation, which promote coarser phases and shrinkage porosities. Furthermore, the high diameter-to-length ratio of the struts limits their ability to bend, making node fracture more likely than ductile bending of the struts.

Considering the concept of brittle deformation behavior of cast lattice structures, the BCC structure was fabricated with the same design, size and alloy using LPBF process in two separate batches to investigate the differences between the production routes. The first batch was produced with low laser power and the second batch with higher laser power and slightly lower scanning speed to improve powder adhesion.

8.3. Additively manufactured lattice structures

The production of lattice structures via direct additive manufacturing, particularly Laser Powder Bed Fusion (LPBF), presents both opportunities and challenges. While LPBF enables the fabrication of highly complex metal geometries, careful

control of the process parameters is essential to ensure structural integrity and performance. Samples that are produced for this study are listed in Table 8-4.

In the first batch of BCC lattice structures produced by LPBF, insufficient laser power resulted in partially fused powder particles remaining adhered to the strut surfaces. These loosely bonded particles detached easily during handling. Metallographic examination revealed a highly porous internal structure caused by incomplete fusion. Furthermore, the micro-porosity distribution was found to be non-uniform, with a greater concentration of voids observed in regions further from the build plate (Figure 8-6). This difference arises from the reduced heat dissipation efficiency at greater distances from the build plate, leading to thermal accumulation. The resulting unstable melt pool behavior increases the risk of porosity and lack-of-fusion defects.

Table 8-4- BCC lattice structures with the overall size of $45 \times 45 \times 45 \text{ mm}^3$ produced by LPBF process out of Al10SiMg alloy, 3 specimens per condition (constant layer thickness of $40 \text{ }\mu\text{m}$ and hatching spacing of $100 \text{ }\mu\text{m}$)

Sample ID	Strut thickness (mm)	Laser power (W)	Scanning speed (mm/s)	Heat treatment
BCC_0.6 – LPBF_1	0.6	200	1200	---
BCC_0.7 – LPBF_1	0.73	200	1200	---
BCC_0.9 – LPBF_1	0.9	200	1200	---
BCC_0.6 – LPBF_2	0.6	250	1100	350 °C for 3 h
BCC_0.7 – LPBF_2	0.73	250	1100	350 °C for 3 h
BCC_0.9 – LPBF_2	0.9	250	1100	350 °C for 3 h

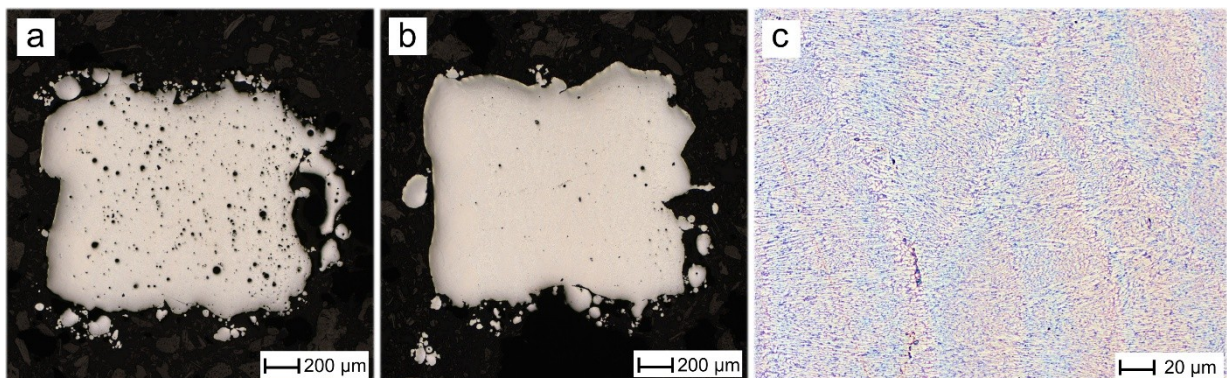


Figure 8-6- Porosity formation inside the strut for the BCC structure produced with LPBF and low laser power a) far from the build plate, b) close to the build plate, c) high magnification of the microstructure in b.

In addition to the observed porosity, the as-printed microstructure was remarkably fine and homogeneous (Figure 8-6c), which is a typical characteristic of LPBF-processed metals. This refinement is primarily due to the extremely high cooling rates associated with the localized and rapid melting/solidification cycles inherent to the LPBF process. Such rapid solidification promotes the formation of a supersaturated solid solution and suppresses the development of coarse dendrites, resulting in a microstructure with ultrafine grains. Furthermore, the repeated heating and remelting of underlying layers promote a characteristic columnar grain growth aligned with the build direction, particularly noticeable in thicker struts.

The fine and homogenous microstructure enhances the strength of the material but can also contribute to residual stresses from the process, which may require post-processing treatments such as stress relief annealing to improve mechanical stability. Without proper stress relief, the ductile deformation behavior typically expected from a refined microstructure may not be realized. Compression testing of these samples revealed that, in the absence of stress-relief treatment and presence of shrinkage porosities, the structures exhibited only elastic deformation, followed by early brittle fracture at locations of shear stress concentration. Figure 8-7 illustrates the average stress-strain curve for the BCC_0.7 lattice structures produced in the first LPBF batch.

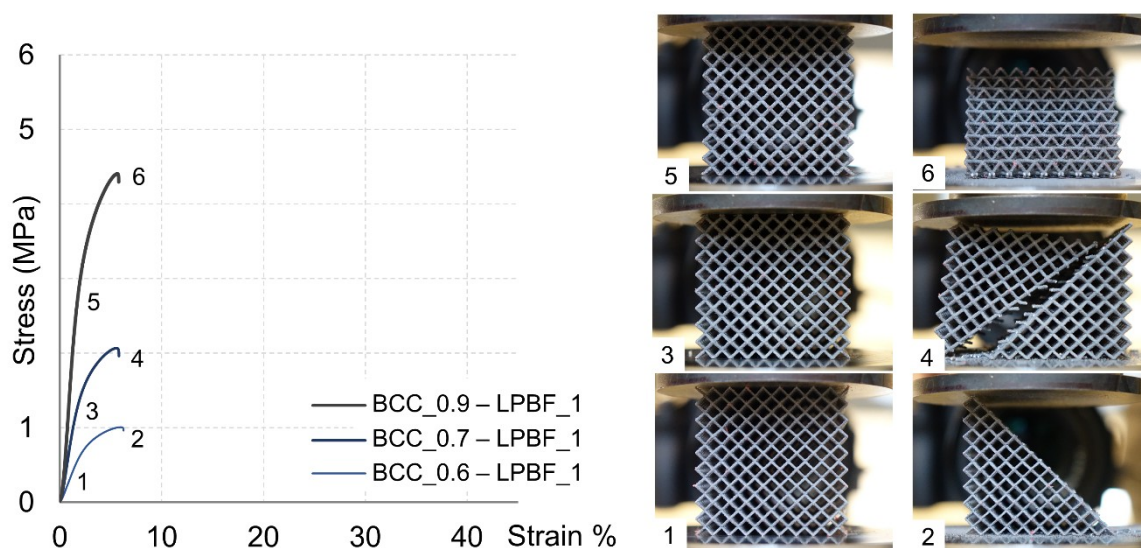


Figure 8-7- Average stress-strain curves of BCC_0.6, BCC_0.7, and BCC_0.9 lattice structure produced by LPBF in the first batch, showing brittle fracture at location of shear stress concentration. Deformation images are shown for one representative sample per condition.

To reduce porosity defects in LPBF-produced lattices, the second batch was fabricated using a higher laser power and a lower scan speed. This adjustment

significantly reduced the number of pores. However, no noticeable change in the microstructure was observed compared to the first batch. In addition, the application of a stress-relief heat treatment altered the mechanical response of the lattice, resulting in fully ductile deformation. The structure was able to compress entirely without any strut fractures. Figure 8-8 presents the average stress-strain curve for the lattices produced in the second batch after heat treatment.

These experiments clearly demonstrate the strong dependence of the mechanical response of LPBF-produced lattice structures on processing parameters. When fabricated with low laser power, the BCC lattices exhibit brittle fracture, failing abruptly despite their high peak strength. This brittle behavior is primarily caused by internal porosity and partially fused regions within the struts, which serve as stress concentrators. However, increasing the laser power and reducing the scan speed significantly reduces porosity, thereby improving structural integrity. Further enhancement is achieved through stress-relief heat treatment, which reduced residual stresses and transformed the deformation behavior from brittle to fully ductile. In this state, the lattice undergoes continuous plastic deformation without any strut fractures, and the presence of a well-defined strain-hardening plateau serves as clear evidence of ductile behavior.

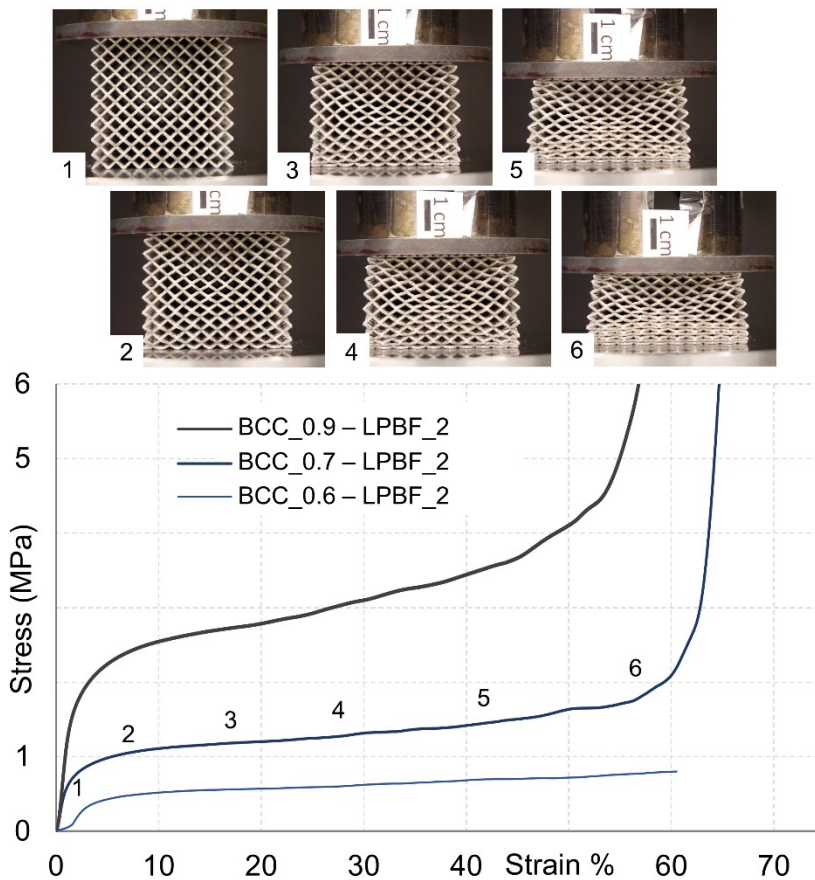


Figure 8-8- Average stress-strain curve of BCC_0.6, BCC_0.7, and BCC0.9 structures produced by LPBF in the second batch after stress-relief annealing. Deformation sequence of one BCC_0.7 lattice sample shown at six different strain levels.

Despite these differences in mechanical performance, certain fundamental deformation characteristics remain consistent across both printed and cast lattice structures. Notably, the formation of shear bands at a 45° angle under compression, which was previously observed in the cast lattices, is also evident in the LPBF-printed counterparts. This suggests that, regardless of the processing route, the intrinsic mechanical response of the BCC lattice topology is influenced by its geometric configuration. However, the ability to tailor the microstructure and porosity through LPBF process optimization and post-processing treatments allows for significantly greater control over the final mechanical properties, making direct additive manufacturing a highly flexible approach for lattice structure production.

Despite this fact, the hybrid investment casting process offers greater flexibility in terms of alloy selection and cost-effectiveness. While LPBF allows precise control over microstructure and porosity, it is often limited to a narrow range of printable alloys and may not be economical for large-scale production. In contrast, investment casting supports a broader selection of alloys, including compositions that are impractical for direct printing. This makes it an excellent technique for fundamental investigations into the deformation behavior of various lattice designs and alloy systems. By initially employing investment casting to explore the structural response and material-specific characteristics, a more informed transition to LPBF can be made. This ensures that only the most promising design–material combinations are selected for direct additive manufacturing.

In the final section that follows, this approach will be extended to investigate the relationship between different alloying concepts and the mechanical performance of cast lattice structures. By analyzing how variations in alloy composition influence deformation behavior, we aim to gain deeper insight into the interplay between material properties and structural geometry. This will support the development of optimized design–material combinations for enhanced structural performance.

8.4. Conclusion: Production and characterization of lattice structures

- Investment casting is a suitable method for producing aluminum lattice structures using cast-resin printed patterns.
- Due to the inherent limitations of the investment casting process (e.g., slow cooling rates), the production of Al-Si alloys results in a coarse microstructure, often accompanied by shrinkage voids, particularly at junction points. These factors count as the main influencing parameters on the deformation behavior of cast lattice structures.

- Even for bending-dominated lattice structures, thick struts do not exhibit visible bending during deformation. Instead, sudden strut fractures occur at the node locations, regardless of the unit cell design.
- The brittle deformation behavior of cast Al-Si lattice structures varies depending on the unit cell geometry: BCC exhibits multiple sudden fractures and pronounced serrations in the stress-strain curve, Diamond: shows higher strength with reduced serration, Kelvin: provides the smoothest deformation behavior and the highest energy absorption efficiency.
- Direct additive manufacturing of lattice structures requires careful tuning of processing parameters and the application of post-processing to achieve ductile mechanical behavior. Otherwise, internal voids, loosely bonded particles, and remained strain result in brittle fracture without any deformation.
- When properly processed, additively manufactured lattice structures can exhibit fully ductile deformation, enabled by a refined microstructure and stress-relief heat treatment.

9. Material and Structural Variations in Cast Lattices

In the previous section, the alloy composition was kept constant, and different production processes were evaluated for manufacturing lattice structures. In this section, the focus shifts to examining the effects of alloying and structural modifications on the mechanical response of cast lattices. The BCC structure was selected as the reference geometry for these investigations, as previous results revealed that it is the most susceptible to the formation of shear bands and exhibits the highest degree of fluctuation in the stress-strain curve. These pronounced instabilities make the BCC structure an ideal candidate for assessing how material refinements and geometrical adjustments can reduce brittle failure and promote more stable, energy-absorbing deformation.

Silicon was selected as the primary alloying element in this study due to its pronounced influence on mechanical properties. By varying the Si content to 3, 7, and 10 wt%, a clear transition from ductile to brittle behavior can be achieved. In addition, two further conditions were examined to assess the effects of microstructural refinement: the addition of strontium (Sr) and casting at a mold temperature of 590 °C to promote a globular grain structure, similar to the stochastic foams discussed in Section 6.3.

To couple these alloying modifications with structural design variations, BCC lattices were produced with two strut diameters: 0.73 mm (BCC_0.7) and 0.9 mm (BCC_0.9). This variation alters the strut slenderness ratio (length-to-diameter), which directly affects the mechanical performance. A higher slenderness ratio (thinner struts) generally increases the risk of bending, resulting in a more compliant lattice with reduced stiffness. Conversely, a lower slenderness (thicker struts) yields a more rigid structure with higher compressive strength but may reduce the lattice's energy absorption capacity.

To further investigate the effect of topology, additional struts with the same diameter were introduced to the BCC_0.7 lattice in two orientations: vertical (BCCz_0.7) and horizontal (BCCx_0.7) as in Figure 5-5. These added elements aim to alter the lattice behavior from bending-dominated to stretch-dominated. In the BCCz_0.7 configuration, the vertical struts primarily bear compressive loads, improving load distribution under compression. In contrast, the horizontal struts in BCCx_0.7 are subjected to tensile forces during node displacement under loading, emphasizing axial stretching. These modifications are intended to evaluate whether the additional struts improve the overall rigidity and alter the mechanical response of the lattice. All the samples produced for this section are listed in Table 9-1.

Table 9-1- Chemical composition and casting condition for the production of cast lattice structures in the advanced study phase (3 specimens per condition)

Alloy	Sample ID	Strut diameter (mm)	Si wt%	Fe wt%	Mn wt%	Mg wt%	Sr wt%	Melt temp (°C)	Mold temp (°C)
AlSi3MgMn	BCC_0.7	0.73							
	BCC_0.9	0.9							
	BCCz_0.7	0.73	3.04	0.08	0.29	0.31	---	731	700
	BCCx_0.7	0.73							
AlSi7MgMn	BCC_0.7	0.73							
	BCC_0.9	0.9							
	BCCz_0.7	0.73	6.98	0.09	0.29	0.29	---	706	700
	BCCx_0.7	0.73							
AlSi7MgMnSr	BCC_0.7	0.73							
	BCC_0.9	0.9	6.96	0.09	0.29	0.30	413	704	700
AlSi7MgMn_Globular	BCC_0.7	0.73							
	BCC_0.9	0.9	6.96	0.09	0.29	0.30	---	705	590
AlSi10MgMn	BCC_0.6	0.6							
	BCC_0.7	0.73							
	BCC_0.9	0.9	10.38	0.10	0.28	0.29	---	683	700
	BCCz_0.7	0.73							
	BCCx_0.7	0.73							

9.1. Ductility adjustment by Si content

As observed in section 8.1.1, the typical deformation behavior of the BCC_0.7 structure made from the AlSi10MgMn alloy, which exhibits brittle characteristics, is primarily governed by the formation of shear bands and sudden strut fracture. This leads to significant fluctuations in the stress-strain curve, resulting in reduced energy absorption efficiency.

To facilitate a meaningful comparison across different lattice structures and alloy compositions, Figure 9-1 presents the average normalized stress-strain curves for BCC_0.7 structures produced from AlSi3MgMn, AlSi7MgMn, and AlSi10MgMn alloys. Normalized stress is used instead of absolute stress values to enable analysis of the mechanical response independent of density. By presenting the results in the form of normalized stress, a clear understanding is gained of how the alloy composition and microstructural changes influence the intrinsic mechanical behavior of the lattice structures.

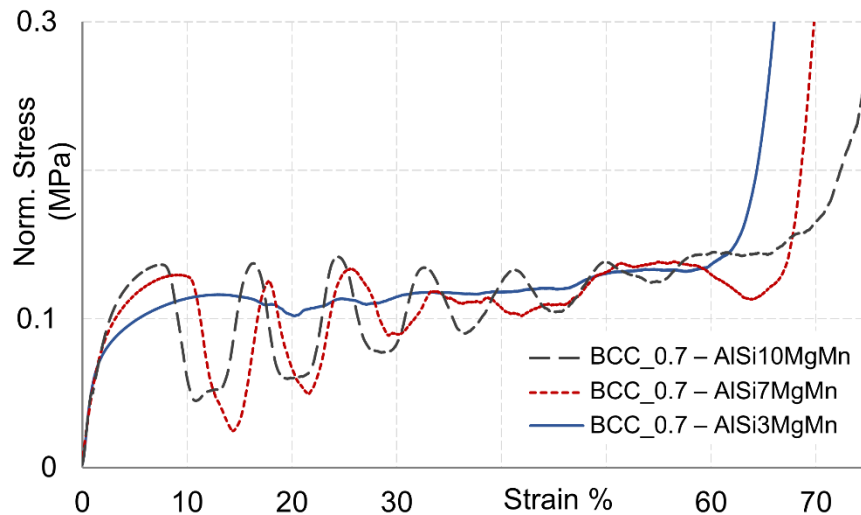


Figure 9-1- Average normalized stress-strain curves of BCC_0.7 lattice structures for aluminum alloys containing different Si contents.

Several key findings emerge from the comparison of BCC_0.7 aluminum lattice structures produced with different Si contents (3, 7, and 10 wt%). For the alloy with 3 wt% Si, the stress-strain curve exhibits a smooth, flat plateau without significant fluctuations. This suggests that reducing brittleness minimizes the likelihood of sudden strut fracture, enabling more homogeneous deformation throughout compression by formation of plastic hinges and strut bending. Unlike additively manufactured lattices, which exhibit significant strain hardening due to their fine microstructure, this cast lattice showed a stable plateau region where stress remained relatively constant as deformation progressed. This behavior indicates that the ductile nature of the alloy allows progressive strut bending and collapse rather than abrupt failure.

As the Si content increases, four main trends emerge, which are presented below together with an explanation of the corresponding mechanisms:

- 1- Higher peak stress and increased brittleness: the increase in overall stress level with higher Si content can be attributed primarily to solid-solution and particle strengthening mechanisms. A higher Si level promotes a greater volume fraction of the Al-Si eutectic and increases the amount of Si dissolved in the α -Al matrix. The resulting Si particles act as barriers to dislocation motion, raising the yield and peak stresses of the alloy. However, this strengthening comes at the expense of ductility. In unmodified Al-Si alloys, the eutectic Si tends to form coarse, plate-like morphologies with sharp edges, which serve as potent stress concentrators. Under compression, these features promote microcrack initiation at lower plastic strains, shifting the failure mode from ductile bending to brittle fracture

of the struts. Consequently, deformation events become abrupt, producing the pronounced stress fluctuations observed in the stress–strain curves of high-Si alloys.

- 2- Decrease in plateau stress: despite the higher intrinsic strength of the base material, the plateau stress decreases with increasing Si content. In ductile lattices, yielding in one strut allows the load to redistribute smoothly to neighboring struts, resulting a stable plateau region. In contrast, brittle fracture events in Si-rich alloys result in the sudden removal of entire load-bearing paths. These abrupt failures not only cause sharp, immediate stress drops but also accelerate cumulative damage, as neighboring struts become overloaded and fail prematurely. The lack of substantial plastic deformation and strain hardening in brittle alloys prevents the maintenance of a steady load-carrying capacity, leading to a lower average plateau stress over the deformation process.
- 3- Initial fracture at lower strains: the earlier onset of fracture in alloys with a high silicon content is directly linked to the reduced fracture strain of the base material. Coarse and angular eutectic Si particles, act as crack nucleation sites. In alloys with low Si content, such as the 3 wt.% Si variant, struts can develop plastic hinges at junctions, enabling progressive and distributed deformation. As the Si content increases, these hinges are suppressed by premature crack formation, causing strut fracture to occur at smaller global strains and shifting the peak stress point to lower strain levels.
- 4- Higher densification strain: the densification strain increases with Si content despite the earlier local failures. This can be explained by the higher residual strength of the intact struts after initial damage, which allows the structure to maintain some load-bearing capacity until larger global strains are reached. Furthermore, fragments of broken struts can interlock or wedge between intact members, generating frictional resistance to compaction. The higher elastic modulus of Si-rich alloys also resists the complete collapse of the structure in the early stages of densification. As a result, although deformation becomes more localized and brittle, the final strain required for full densification is greater in high-Si lattices.

The same key trends observed in the BCC_0.7 lattices—peak stress increase, plateau stress decrease, fracture at lower strains, and higher densification strain with increasing Si content—were also evident in the BCC_0.9 structures as Figure 9-2. However, due to the larger strut diameter, BCC_0.9 exhibited higher stress levels across all alloys, as the thicker struts could tolerate greater loads before failure.

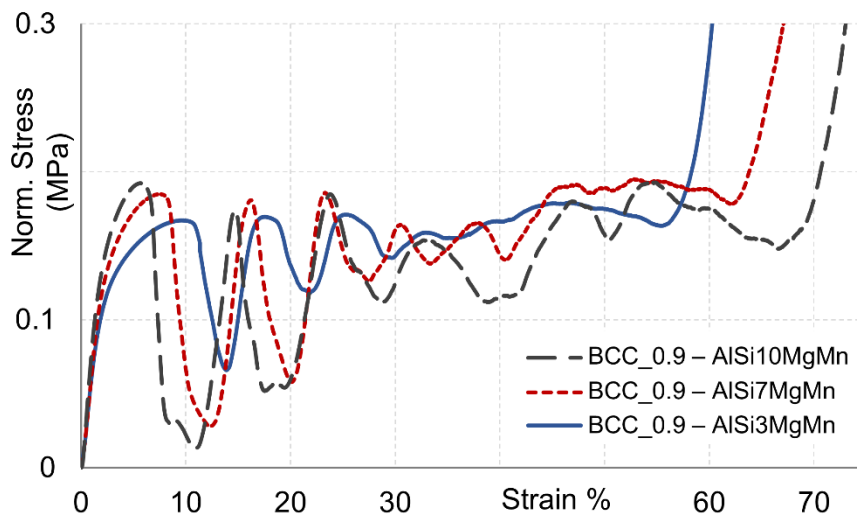


Figure 9-2- Average normalized stress-strain curves of BCC_0.9 lattice structures for aluminum alloys containing different Si contents.

When the strut diameter increased from 0.7 to 0.9 mm, even the most ductile alloy with 3 wt% Si no longer exhibited homogeneous deformation. Instead, the deformation behavior shifted to brittle fracture, producing stress serrations in the compression curve. This change can be explained by the reduced slenderness ratio of the thicker struts. With a lower length-to-diameter ratio, the struts experience a higher fraction of direct axial loading and a reduced capacity to form plastic hinges. In the thinner BCC_0.7 structure, these hinges enable progressive, distributed deformation, leading to a smooth, plateau-like stress response for 3 wt% Si. In contrast, in the thicker BCC_0.9 struts, the increased stiffness and higher hinge formation tolerance suppress this mechanism. As a result, localized failures become more abrupt and pronounced, resembling the brittle response of high-Si alloys.

The geometric effect is further amplified by microstructural and defect-related factors inherent to the casting process. The larger volume of each strut also statistically increases the probability of containing critical casting defects such as porosity, particularly near nodes that solidify last. Under predominantly axial loading, these flaws experience higher stress level and act as efficient crack initiation sites. Furthermore, the increased cross-section at the nodes raises local constraint, limiting the potential for ductile bending. Together, the geometric shift toward axial loading, the more brittle microstructure, the higher defect probability, and the increased junction limitation explain why BCC_0.9 lattices deform through brittle, serrated fracture even in alloys that behave ductile when the struts are thinner.

9.2. Effect of Sr addition

To investigate the influence of second phase refinement on the mechanical properties of lattice structures, Sr was added to the AlSi7MgMn alloy to modify the morphology of the eutectic Si phase and refine the Si particles. However, Sr addition did not lead to any measurable improvement in the mechanical properties or deformation behavior of the BCC_0.7 and BCC_0.9 lattices. As shown in Figure 9-3, the stress-strain curves of samples with and without Sr are nearly identical, suggesting that either the Sr-induced microstructure changes had a negligible influence on the overall mechanical response, or Sr did not effectively contribute to the modification of the Si phase.

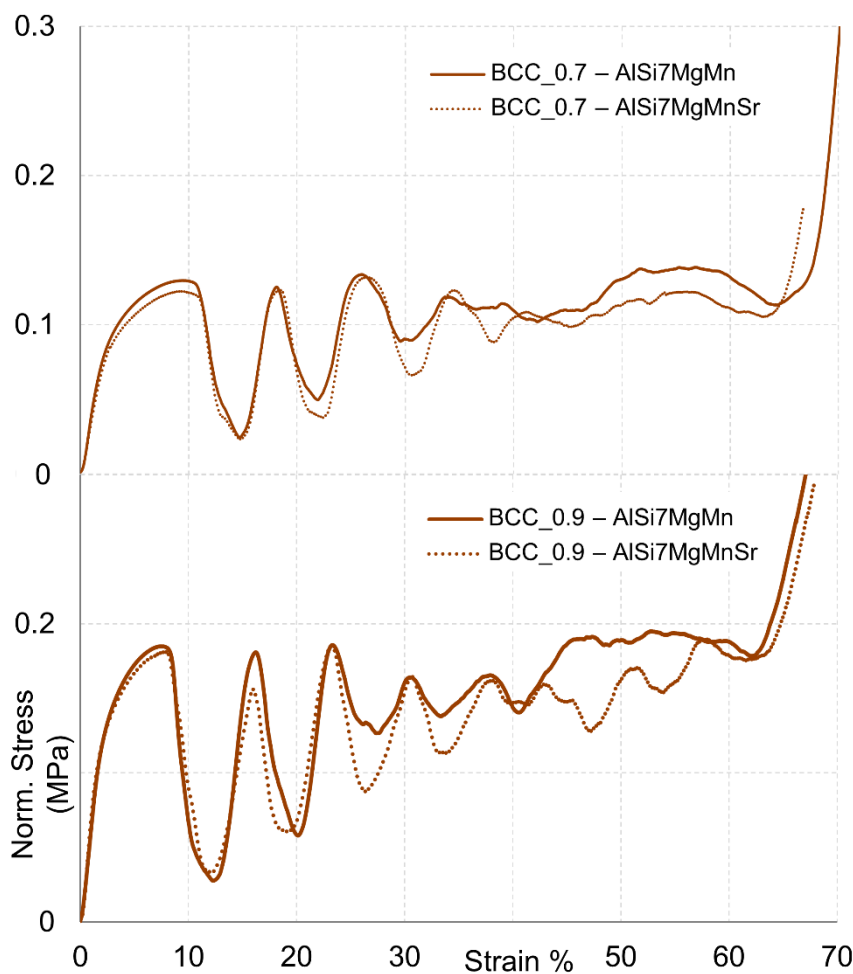


Figure 9-3- Average normalized stress-strain curves of BCC_0.7 and BCC_0.9 lattice structures for AlSi7MgMn base alloy with and without Sr addition.

One possible explanation for this outcome is the relatively large strut diameters in the tested lattices compared to the thinner struts previously examined in stochastic foams. In foam structures with thin struts, deformation is often bending-dominated, making them more sensitive to ductility enhancements in the base material. In contrast, the deformation of thick-strut lattices is governed more by localized stress concentrations and crack initiation at weak points, such as casting defects and high-Si regions. Since Sr modification improves eutectic morphology but does not

eliminate shrinkage porosity or casting defects in cast lattices effectively, its influence on ductility and overall structural performance remains limited in this context.

Another critical factor to consider is the non-uniform modification of the silicon phase, influenced by variations in cooling rate during block mold casting. As noted in Section 7.1.1, even a slight change in cooling rate can significantly impact the modification of Si particles in aluminum foams. In regions of the foam structure closer to the flask wall, where cooling rate was higher, strontium was effective in refining the Si morphology. In contrast, areas further from the flask wall, where cooling was slower, exhibited little to no modification, and the Si particles retained their coarse morphology. A similar trend was observed in the lattice structures studied here. Microstructural analysis revealed that struts near the outer sections (first node row, close to the flask wall) displayed well-modified Si particles (Figure 9-4b), whereas most struts in the central region showed minimal modification (Figure 9-4c). Figure 9-4 illustrates this non-uniform Sr modification across the BCC_0.9 lattice.

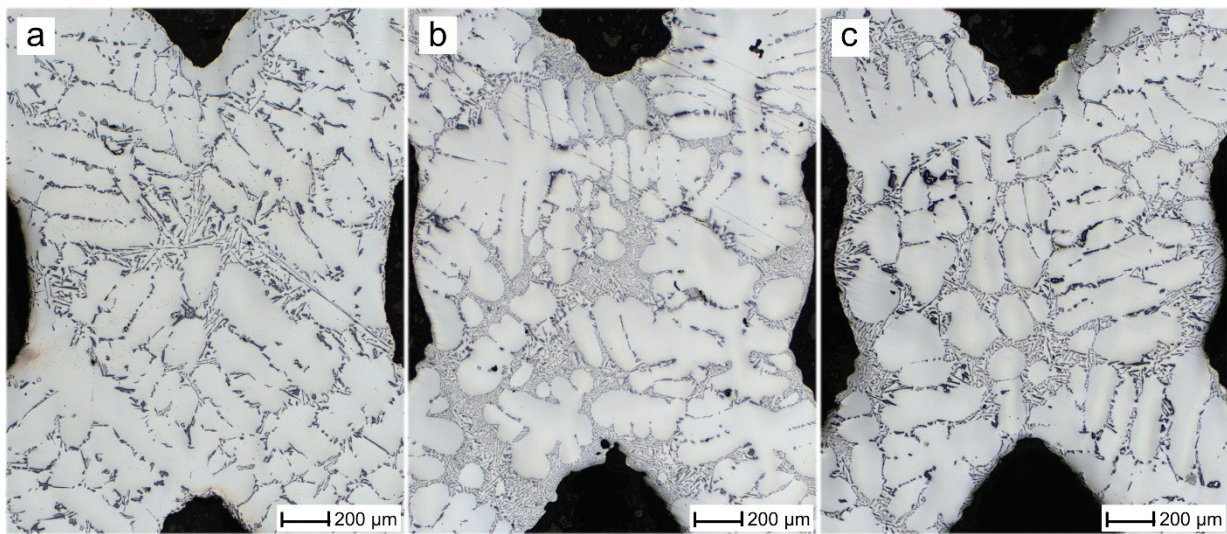


Figure 9-4- Microstructure of BCC_0.9 lattice structures: a) Al₇SiMgMn alloy at the first node row close to the flask wall, b) Al₇SiMgMnSr alloy at the first node row close to the flask wall, and c) Al₇SiMgMnSr alloy at the third node row closer to the middle of the lattice.

Given that most of the load-bearing capacity in the lattice structure arises from the interconnected struts in the middle section, the localized modification of Si particles in the outer struts has little to no effect on the overall mechanical properties. Since these outer struts constitute only a small fraction of the total structure, any potential improvement in their ductility does not significantly influence the global deformation behavior or enhance the structural performance.

These findings emphasize the complexity of applying microstructural modifications to cast lattice structures and highlight the critical roles that both geometric factors and solidification conditions play in determining mechanical performance. While Sr addition proves effective in cast aluminum foams under favorable cooling condition closer to the flask wall, its impact on lattice structures remains limited when strut dimensions are relatively larger and cooling rates vary across the casting. As shown in Figure 9-4, the modification of the silicon phase by Sr is non-uniform: struts near the flask wall, experiencing higher cooling rates, exhibit refined Si morphology, whereas struts located centrally, with slower cooling, retain coarse and unmodified silicon particles.

9.3. Effect of globular grain structure

To further investigate the role of microstructural fine-tuning in lattice structures, BCC_0.7 and BCC_0.9 specimens were produced by casting the AlSi7MgMn alloy in molds preheated to 590 °C. This approach was chosen to replicate the findings reported for stochastic foams in section 6.3.1 and to promote the formation of a globular aluminum grain structure. The resulting globular grain structure obtained for the BCC_0.9 lattice is compared in Figure 9-5 to the dendritic grain structure for the same lattice produced when casting in molds preheated to 700 °C.

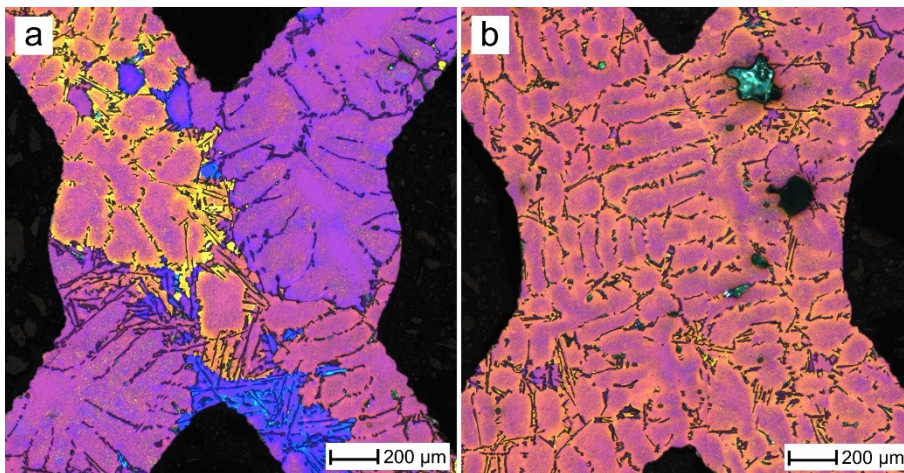


Figure 9-5- Microstructure of BCC_0.9 lattice structures made from Al7SiMgMn alloy cast in molds preheated to a) 590 °C and b) 700 °C.

Consistent with the observations for stochastic foams, casting lattice structures in cooler molds promotes the formation of a globular, multi-grain aluminum microstructure rather than the predominantly single, dendritic grains typically found within each strut. In the case of molds preheated to 590 °C, the faster cooling rate restrict dendrite growth, encouraging the nucleation of multiple grains. By contrast, casting in molds preheated to 700 °C results in slower cooling, allowing dendrites to grow across the entire strut diameter, producing a coarse, single-grain

structure. However, similar to the effect of Sr modification, this microstructural alteration did not lead to any improvement in deformation behavior or key mechanical characteristics as in Figure 9-6.

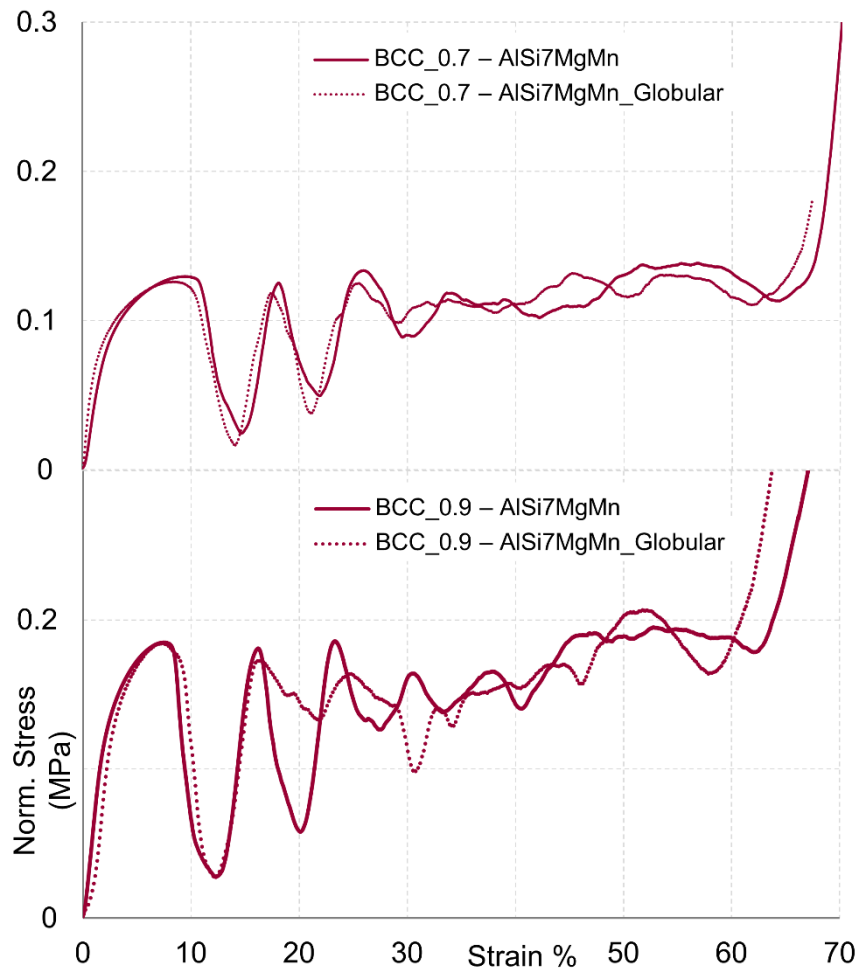


Figure 9-6- Average normalized stress-strain curves of BCC_0.7 and BCC_0.9 lattice structures for dendritic (normal) and globular aluminum grain structure of AlSi7MgMn base alloy.

These findings further emphasize that microstructural improvements, whether through Si phase refinement or changes in the solidification morphology of the aluminum matrix, do not necessarily lead to improved mechanical properties for lattice structures. In bulk castings, the transition from a dendritic to a globular grain structure is known to enhance ductility and refine mechanical responses by reducing stress concentrations and promoting uniform plastic deformation. However, in lattice structures with relatively thick struts, such benefits appear negligible.

Similar to the Sr modification of the eutectic Si phase, altering the grain morphology of the aluminum matrix did not lead to measurable improvements in peak strength, plateau stress, or deformation behavior for either BCC_0.7 or BCC_0.9. This suggests that the dominant factors governing failure in these structures are related more to strut geometry, stress distribution, and the presence of casting defects rather than

intrinsic material ductility. In particular, due to the larger strut diameters, deformation in these lattices is largely governed by localized stress concentrations rather than by gradual, ductility-controlled plastic flow.

9.3.1. Discussion on the effect microstructural changes in lattice structures

The influence of microstructural refinement in cast lattice structures is markedly different from that in bulk materials or even stochastic foams. While reductions in eutectic phase content, as observed in the case of BCC_0.7, could promote more ductile deformation through increased strut bending and hinge formation, this effect was easily suppressed by increase of Si and strut diameter.

In lattices with inherently brittle deformation modes (e.g., BCC-type designs), stress is concentrated at node junctions, causing localized fracture to dominate over bending. Under these conditions, microstructural refinement alone, even through chemical modification or grain structure change, cannot fully overcome the structural limitations. This contrasts with foams, where stress is more evenly distributed and strut bending is possible. Therefore, the microstructural improvements in foams can be transferred more directly into improved ductility.

The potential benefits of refining the microstructure are further diminished by casting-specific limitations. Due to the lack of effective feeding paths in lattice struts, shrinkage porosity and other casting defects are unavoidable. Although the BCC_0.7 lattice contains smaller defects than BCC_0.9 because of smaller struts and less material accumulation at nodes, the overall deformation behavior remained similar. This indicates that defect reduction alone is insufficient when the structural design itself promotes stress concentration.

These findings highlight that optimizing mechanical performance in cast lattice structures requires an integrated approach. Microstructural control through alloy composition and modification should be paired with mesostructural and macrostructural optimization, like strut slenderness and cell geometry, if the benefits of improved alloy ductility are to be fully realized.

9.4. Slenderness of struts

In addition to microstructural modifications, the impact of geometric changes, such as slenderness and the addition of extra struts, on lattice structure deformation was further investigated. In general, slenderness plays a critical role in determining the deformation mechanism of lattice structures. When struts are more slender (higher length-to-diameter ratio), they are more prone to bending under compressive loading, which can enhance ductility and energy absorption by allowing gradual and distributed deformation rather than abrupt fracture. In contrast, as strut thickness

increases, the deformation mode changes from bending-dominated to more stretch-dominated behavior, where the struts bear higher direct axial loads. This shift typically increases the overall strength of the structure but also makes it more susceptible to sudden failure, especially in brittle materials.

To further investigate the effect of slenderness on deformation behavior, additional BCC lattice structures with strut diameter of 0.6 mm (BCC_0.6) were further cast along BCC_0.7 and BCC_0.9 using the most brittle alloy (AlSi10MgMn). As shown in Figure 9-7, the ductility and homogeneous deformation in the plateau regime were highest for the BCC_0.6 structure, although the tolerable load is lower. This confirms that decreasing the strut diameter promotes a more bending-dominated response, enhancing ductility even in a brittle alloy. Therefore, ductility can be tailored not only through chemical composition, e.g. Si content, but also by adjusting the geometric parameters of the lattice.

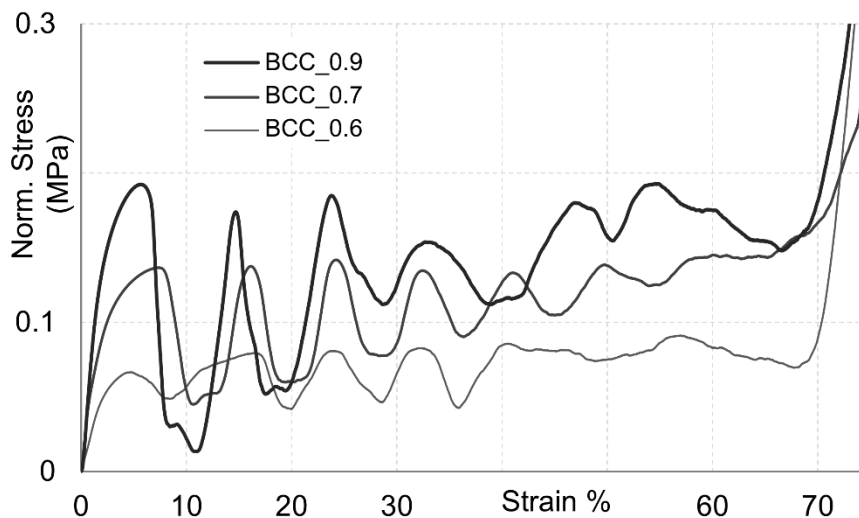


Figure 9-7- Average normalized stress-strain curves of BCC lattice structure with strut diameters of 0.6, 0.73, and 0.9 (BCC_0.6, BCC_0.7, and BCC_0.9) cast by AlSi10MgMn alloy.

9.5. Stretch dominated structure

The addition of extra struts to the BCC lattice structure fundamentally alters the load-bearing characteristics and deformation response. In this study, two modifications were introduced to the BCC_0.7 structure: BCCz_0.7 that extra struts were added in the vertical direction, aligned with the compression axis and BCCx_0.7 that extra struts were added in the horizontal direction, perpendicular to the compression axis as presented in Figure 5-5.

The introduction of additional vertical struts in BCCz_0.7 had a profound impact on mechanical performance. Since these extra struts directly aligned with the loading

direction, they contributed significantly to load-bearing capacity, effectively transforming the structure from a predominantly bending-dominated to a stretch-dominated configuration. As a result, for all three alloys (AlSi3MgMn, AlSi7MgMn, and AlSi10MgMn), the peak normalized strength increased substantially, reaching up to five times the values observed in the original BCC_0.7 structure (Figure 9-8). This enhancement is attributed to the additional load paths provided by the vertical struts, which increased structural stiffness and reduced localized bending effects. Consequently, the deformation behavior became more stable, with a reduced tendency for premature failure.

Beyond the initial strengthening effect of the load bearing strut, the deformation response varies depending on the alloy's ductility. For the most ductile alloy (AlSi3MgMn), deformation remains homogeneous even after the peak stress, similar to what was observed in the standard BCC_0.7 structure. After reaching the peak stress, the stress experiences only a minor drop before stabilizing into a smooth plateau, indicating stable load-bearing behavior. As the Si content increases, two key effects emerge:

1. Greater post-peak instability: while the peak strength continues to rise with higher Si content, the stress drop after the peak becomes more pronounced. This suggests that brittle alloys lack the stability to maintain a controlled deformation response. At later deformation stages, this instability results in greater stress fluctuations, which further emphasizes the brittle nature of the material.
2. Decrease in plateau stress: this trend, also observed in the standard BCC_0.7 structure, is even more pronounced in BCCz_0.7 due to the dominant load-bearing role of the additional vertical struts. Since these extra struts carry most of the load, the increase in peak stress with higher Si content is less significant compared to the standard BCC_0.7 structure. However, because the brittle alloys (Al-7Si and Al-10Si) experience sharp stress drops after the peak, their average stress over the deformation process is lower. In contrast, for the most ductile alloy (Al-3Si), the stress drop after peak is minimal, allowing the plateau stress to remain remarkably high. This emphasizes that while higher Si content strengthens the material, it also reduces the ability of the lattice to sustain a high plateau stress due to instability and localized failures.

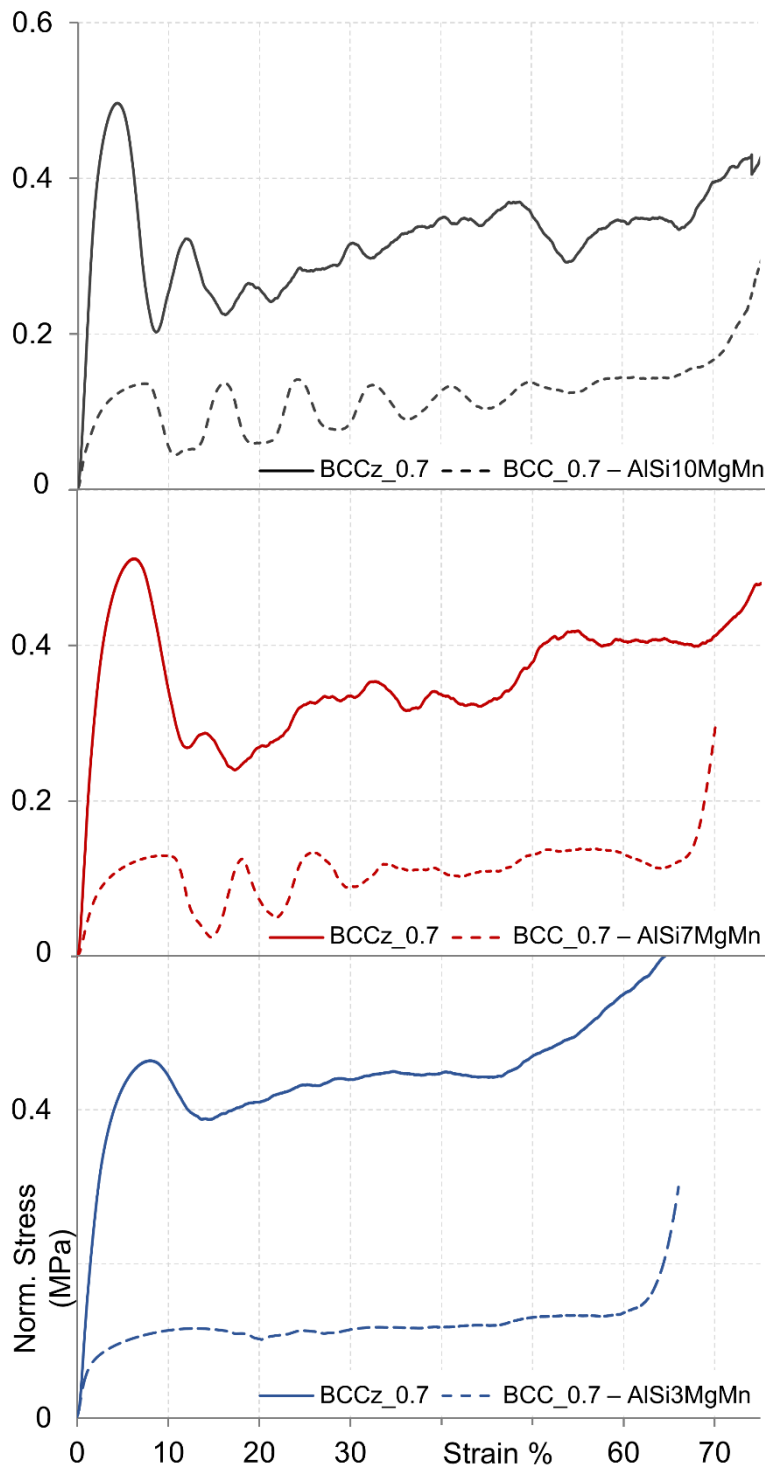


Figure 9-8- Average normalized stress-strain curves of BCC lattice structure (D: 0.73 mm) with extra strut in the vertical direction and without extra strut for AlSi3MgMn, AlSi7MgMn, and AlSi10MgMn alloys.

Adding an additional horizontal strut to the BCC_0.7 structure (creating BCCx_0.7) does not influence the deformation behavior or mechanical properties of the lattice, regardless of the alloy composition. The stress-strain curves for BCCx_0.7 closely resemble those of the original BCC_0.7 structure, showing no noticeable differences in peak stress, plateau stress, or overall deformation characteristics (Figure 9-9).

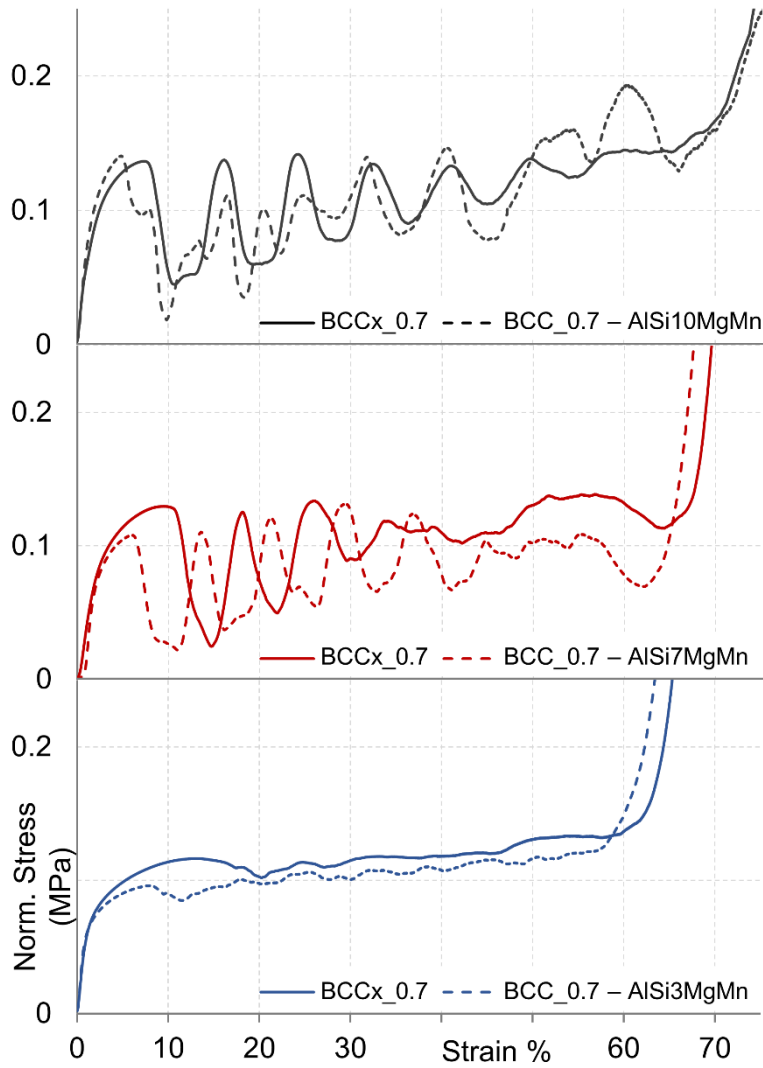


Figure 9-9- Average normalized stress-strain curves of BCC lattice structure (D: 0.73 mm) with extra strut in the horizontal direction and without extra strut: AlSi3MgMn, AlSi7MgMn, and AlSi10MgMn alloys.

The primary reason for this behavior lies in the nature of the loading experienced by these extra struts. In BCCx_0.7, the horizontal struts are oriented perpendicular to the loading direction, meaning they are primarily subjected to tensile forces during node displacement and cell deformation rather than compressive loads. However, due to the inherent casting defects and the coarse microstructure, these struts lack the necessary integrity to effectively carry tensile loads. As a result, they fail prematurely or do not significantly engage in the deformation process, making them mechanically ineffective.

Furthermore, the casting-induced porosities and the brittle behavior of Al-Si alloys intensified this issue. In tension, defects such as pores act as stress concentrators, promoting early failure before the struts can contribute meaningfully to the structural response. Unlike in compression, where the material can still provide some resistance, tensile failure is immediate when flaws are present. This explains why no improvement in mechanical performance was observed for BCCx_0.7

despite the addition of extra material. It proves that the effectiveness of lattice design modifications depends not only on the geometry, but also on how the additional elements interact with the applied load.

9.6. Conclusion: Effectiveness of changes in lattice structures

- Variations in alloy ductility were targeted by adjusting the silicon content of the Al-Si alloy. Increasing the Si content, which reduced the alloy's ductility, led to the several trends in BCC lattices with strut diameters of 0.73 mm and 0.9 mm: increase of peak stress, decrease of plateau stress, decrease of strain at peak stress, and increase of densification strain.
- Sr addition modified the morphology of silicon particles only in regions close to the flask wall (outer structure region), while most of the internal lattice remained unmodified. Consequently, Sr addition did not produce any measurable changes in deformation behavior or mechanical properties.
- Casting in molds preheated to 590 °C produced a globular, multi-grain aluminum microstructure, consistent with previous observations for stochastic foams. However, this refinement in grain morphology did not lead to noticeable changes in the mechanical performance of the lattice structures.
- The diameter and slenderness of the struts are key structural parameters influencing the deformation behavior of lattice structures. Reducing the strut diameter generally increases ductility at the expense of allowable strength. In slender struts, the mechanical response more closely reflected the intrinsic properties of the base material, meaning that if the alloy is ductile, its ductility will be more clearly expressed in the structural behavior. Conversely, thicker struts tend to suppress the apparent ductility, even when the base material itself is ductile, leading to a more brittle-like deformation response.
- Incorporating additional load-bearing elements, such as struts aligned with the loading direction, increased the tolerable strength of the lattice. The intrinsic behavior of the base material influenced both the peak stress and the post-peak deformation response in this condition. Increasing the Si content raised the peak stress moderately but had a pronounced negative impact on post-peak performance, leading to larger stress drops and more brittle deformation. This indicates that while higher Si content enhances material strength, it promotes instability and localized failure.
- Adding struts perpendicular to the loading direction did not alter the overall deformation behavior of the lattice. These elements were primarily subjected to tensile stresses and, likely due to inherent casting defects, failed prematurely without significantly contributing to the load-bearing capacity or deformation process.

10. Evaluation of Influencing Factors Across Hierarchical Levels in Cellular Structures

Considering the range of modifications applied at different hierarchical levels in this study, distinct influences were observed for each cast cellular structure type, namely stochastic foams and regular lattice structures. These modifications spanned microstructural features (alloy chemistry and heat treatment), mesostructural factors (strut orientation, and slenderness), and a macrostructural parameter (unit cell design).

In stochastic foams, the random cell arrangement that is surrounded by thin struts promote a predominantly bending-dominated deformation mode. Plastic hinges can readily form at node locations, enabling struts to deflect into S- or C-shaped curves before failure. Under these conditions, microstructural characteristics exert a strong influence on the macroscopic response. Alterations such as eutectic modification, intermetallic phase refinement, or precipitation hardening directly influence the local strut failure mechanisms, which in turn control the foam's overall deformation behavior. Since the foam contains a statistically large number of struts with diverse load paths, improvements or degradations in local ductility are effectively averaged across the structure, resulting in a relatively direct and predictable microstructure–property relationship.

The irregular cell geometry of stochastic foams promotes stress concentration in locally weaker regions, making the structure more sensitive to variations in solidification morphology or phase distribution. Change in the solidification microstructure of primary grains were achieved here by promoting exogenous solidification from the strut walls inward. When instead of a single aluminum dendrite, a multi-grain primary aluminum morphology was obtained, combined with a homogeneous silicon phase distribution, crack propagation paths were deflected and arrested more effectively, leading to a measurable gain in ductility. Eutectic modification further reduced local stress concentrations at brittle particle sites, delaying crack initiation and enabling more uniform plastic deformation. The most significant mechanical improvements, however, were achieved when strengthening was governed by fine in-situ precipitates formed through heat treatment. In this case, the second-phase particle size and distribution could be tuned to provide both high strength and stable plastic flow, mimicking the strengthening mechanisms typically observed in bulk materials.

In contrast, regular lattice structures distribute loads along well-defined load paths, which can partially mask the influence of microstructural refinement. Their

deformation behavior is often governed by structural stability, resistance to buckling, and the presence of casting defects. Therefore, the intrinsic properties of the base material only become decisive when strut slenderness is high or cross-sectional dimensions are small enough for bending to replace compressive crushing as the dominant mode.

This behavior is strongly influenced by the manufacturing route. In cast lattices, unavoidable solidification shrinkage act as stress concentrators, accelerating local fracture initiation, especially in designs like BCC that are already prone to localized failure. In contrast, lattice structures produced by laser powder bed fusion (LPBF) exhibit an ultra-fine microstructure with minimal porosity, allowing even geometries most susceptible to brittle deformation in cast form to undergo fully ductile, stable deformation without strut fracture. This comparison highlights that the apparent suppression of microstructural effects in cast lattices is not a fundamental property of the geometry itself, but rather a consequence of defect population and the microstructural coarseness dependent on the casting process. By eliminating these casting-related limitations, as in LPBF, microstructural refinement can indeed dominate deformation behavior, even in the worst lattice designs.

While thicker struts can increase load-bearing capacity, they often promote a more brittle, fracture-driven failure and reduce the structure's energy absorption efficiency. In contrast, thinner struts not only allow for bending-dominated deformation but also make it easier for microstructural adjustments to influence mechanical performance. However, in the hybrid investment casting of lattice structures, the production of sacrificial patterns imposes significant limitations on reducing strut thickness. Thin-strut patterns frequently fail during printing due to insufficient stiffness, leading to distortion or collapse before casting. Consequently, the high slenderness of structural elements, desirable for improving ductility, cannot always be implemented. This restriction means that direct optimization of mechanical properties through thinner struts is still largely constrained by the limitations of pattern production.

Several mesostructural factors can be adjusted to complement and enhance the effects of microstructural improvements. These include optimizing strut orientation or introducing additional structural elements that help distribute stresses more evenly. In this study, incorporating improved alloy concept into a lattice design modified with an extra compression load-bearing strut demonstrated that high strength can be achieved while simultaneously improving ductility. This highlights

that combining improvements across different hierarchical levels can produce a tailored mechanical response with superior overall performance.

Adjusting slenderness ratio within the limits of pattern production can also help shift the failure mode away from fracture-dominated behavior toward more stable, progressive deformation as observed for the BCC structure with struts of 0.6 mm. Increased slenderness promotes bending and plastic hinge formation, which lowers stress triaxiality and reduces the probability and driving force for defect-controlled fracture. This enables load redistribution and shifts deformation to stable, progressive deformation.

Modifying the unit cell topology as a macrostructural parameter introduces fundamental changes in the mechanical response, even when relative density is kept constant. Different topologies distribute loads through bending, stretching, or mixed modes, and thereby define the dominant deformation and failure mechanisms. For example, Kelvin structures promote a more uniform load distribution and progressive collapse, while BCC structures are inherently more prone to localized failure due to stress concentration at nodes. In this sense, topology acts as a framework that either amplifies or suppresses the effects of microstructure and strut geometry. A ductile alloy or refined microstructure can only fully express its benefits in a topology that avoids premature fracture, while slenderness or orientation adjustments at the mesostructural level are often most effective when supported by a topology with stable load paths. Current optimization approaches in lattice design, often rely on simulation-based studies to identify unit cell topologies that maximize stiffness, energy absorption, or toughness at a given density. Ultimately, the unit cell topology determines the stage on which microstructural and mesostructural modifications interact, making it the most powerful tool to tailor the balance between strength, ductility, and energy absorption in cellular metals.

As a whole, microstructural refinement, mesostructural tuning, and macrostructural design act in synergy, where microstructure dictates the local failure mechanisms, mesostructure adjusts the stability of load-bearing elements, and the unit cell topology sets the overall deformation mode. Only through the interaction of all three levels can cellular metals be truly optimized in terms of strength, ductility, and energy absorption.

11. Conclusion and Outlook

In the current dissertation, the influence of parameters across the micro, meso, and macro scales has been investigated to fulfill the proposed hypothesis.:

- The casting process plays a decisive role in microstructural development, particularly in the formation of the primary solidification microstructure. Certain process constraints, such as a narrow cooling rate range, limit the attainable microstructures. However, when combined with the small solidification volumes imposed by strut geometries, foam casting exhibits higher sensitivity to microstructural changes achievable through cooling rate adjustments, such as silicon phase modification and intermetallic morphology alterations.
- Changing the alloy's strengthening concept from a eutectic second phase (Al-Si system) to nano in-situ precipitates produced by aging (Al-Cu system) fundamentally alters foam deformation behavior. The transition to Al-Cu alloys, combined with optimized heat treatment, not only increases the normalized strength but also eliminates brittle and unpredictable plateau-stage deformation.
- Direct additive manufacturing of metallic lattice structures requires tight process control and is constrained by the limited range of commercially available materials. This restricts the scope for alloy and structural optimization. Hybrid investment casting, on the other hand, offers a feasible alternative for exploring alloy-structure interactions that can later be transferred to additive manufacturing methods.
- Compared with stochastic foams, lattice structures show less effects from microstructural changes. Structural modifications, however, exert a stronger and more direct influence on mechanical properties and deformation behavior.
- Coarse cast microstructure and casting defects in thick struts of lattice structures limit their ability to bend, leading to fracture before the expected ductile deformation occurs. Reducing strut diameter and increasing slenderness make the relative effect of mechanical characteristics of the base material more evident.
- Using high-strength alloy concepts increases initial strength parameters such as peak stress but reduces post-peak stability. In these alloys, stress drops and fluctuations during the plateau stage become more pronounced, lowering both plateau stress and the energy absorption efficiency.

Building upon the findings of this research, several promising avenues can be explored to further advance the understanding and application of aluminum-based cellular structures:

- In-situ monitoring and modeling of solidification behavior in micrometer-scale struts to better capture real-time morphological evolution: this could enhance predictive capabilities and support the design of more robust microstructures.
- Heat-Treatable Al-Cu Alloys in lattice structures: investigating the application of heat-treatable aluminum alloys, such as the Al-Cu system, in the fabrication of lattice structures could provide valuable insights. By examining how microstructural changes induced by heat treatment interact with various cell designs, one can assess the potential for enhancing mechanical properties.
- Extending the microstructural tuning strategies to other alloy systems: such as Al-Mg or Al-Zn, which may offer different strengthening mechanisms and deformation behaviors in lattice structures.
- Hybrid structures that combine different unit cell types or materials could open new opportunities for locally tailored properties, particularly for applications requiring both stiffness and energy absorption.
- Assessment of long-term performance: evaluating the long-term mechanical performance and cyclic loading behavior of cast aluminum lattice structures under various service conditions is essential for their application in industries. Research into the effects of prolonged exposure to different temperatures and environments can provide insights into the durability and reliability of these materials.
- Studying various lattice designs for thermal management: designing lattice structures optimized for use as heat sinks could combine the benefits of lightweight architecture with effective thermal dissipation, potentially serving high-performance applications in electronics or automotive components.

12. References

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13. Publications

- M. Firoozbakht, A. Bührig-Polaczek, Hybrides Feingussverfahren für Gitterstrukturen: grundlegende, umfassende und zuverlässige Methode, Giesserei Forschung 11.2025.
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<https://doi.org/10.1002/adem.202400158>
- M. Firoozbakht, A. Bührig-Polaczek, Proactive, Innovative and Flexible Foundry Institute – Review of Investment Casting Researches (Invited Plenary Talk), 63rd IFC, Portorož, Slovenia, 13-15 Sep. 2023.
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- M. Firoozbakht, A. Blond, A. Bührig-Polaczek, C. Fleck, A.C. Kaya, Microstructure Improvement of Al-7Si Alloy Open-Pore Foams by Tuning of Investment Casting Process, CellMat 2022, Dresden & Online, Germany, 12-14 Oct. 2022.
- M. Firoozbakht, A. Blond, A. Bührig-Polaczek, C. Fleck, A.C. Kaya, Modification of Iron Intermetallic from Beta Phase to Chinese Script Phases in the Ultrathin Cast Aluminum Foams, MSE 2022 (Materials Science and Engineering Congress), TU Darmstadt, Germany, 27-29 Sep. 2022.

14. CV

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07.2014 – 09.2017	Tabarestan Steel Foundry, Qaem Shahr, Iran Senior engineer in R&D
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15. Abstract English

This study investigates the fabrication, microstructural evolution, and mechanical performance of open-cell aluminum foams and lattice structures produced by investment casting. Aluminum cellular materials are of significant interest for lightweight structural applications due to their high specific strength and energy absorption capacity. However, the combined effects of casting conditions, alloy composition, and architectural design on the deformation behavior of such structures remain insufficiently understood. By integrating the effect of alloy alteration with structural design, this work establishes new insights into the process–structure–property relationships of cast cellular aluminum materials.

In Al-Si alloy foams, the mold temperature was found to exert a dominant influence on solidification microstructure and deformation behavior. Higher mold temperatures promoted dendritic microstructures with coarse eutectic silicon phases localized at the strut surfaces, which resulted in brittle deformation, premature failure, and serrated stress–strain responses. In contrast, lower mold temperatures favored globular aluminum grains and a more homogeneous silicon distribution, enabling ductile deformation and smoother stress–strain curves. These benefits, however, were offset by processing challenges, as insufficient mold filling and thinner struts limited the properties improvement of foams produced under such conditions.

Local cooling rate variations during foam casting further influenced finer microstructural features. Although the overall change in cooling rate was moderate, faster cooling enhanced the effectiveness of strontium modification, refined the eutectic phase, and altered the morphology of iron-rich intermetallics. These findings highlight that sensitivity to microstructural change is higher when solidification happens at a slow cooling rate and in a confined solidification space, where local diffusion and solidification dynamics strongly affect phase formation.

The deformation behavior was also shown to be highly sensitive to the size and distribution of second-phase particles within micrometer-sized struts. Coarse, brittle phases such as eutectic silicon led to strain localization and premature failure, while refinement of these phases or their substitution by fine precipitates, as obtained in heat-treated Al-Cu alloys, promoted ductile and homogeneous deformation. In foams with a uniform and refined microstructure, the plateau stress correlated directly with intrinsic material properties such as strut hardness, reflecting similar trends to those observed in bulk alloys. This demonstrates that when brittle constituents are minimized, the foam-level mechanical response is governed by fundamental material parameters.

Lattice structures produced by indirect additive manufacturing enabled systematic variation of cell geometry, which provided further insight into the interaction between architecture and material behavior. The results showed that relative density is less dominant in structures with thicker struts, where deformation is primarily governed by cell design. In bending-dominated lattices, brittle failure without bending could occur when unfavorable alloy chemistry or casting defects were present, indicating that architectural advantages can be neutralized by poor microstructural quality. At thicker strut dimensions, fine-scale microstructural modifications had little influence on deformation. However, as struts became thinner, microstructural sensitivity increased markedly, revealing an interaction between geometry and material behavior. Moreover, tensile load-bearing elements in cast lattice structures were shown to be mechanically ineffective, suggesting that such features can be avoided in design to improve efficiency.

Overall, this work clarifies how processing, microstructure, and geometry interact to determine the performance of cast aluminum foams and lattices. By linking alloy design to structural behavior across multiple length scales, the study provides a comprehensive framework for tailoring lightweight cellular metals through combined microstructural and geometrical optimization. The findings contribute to advancing the understanding of cast cellular aluminum materials and offer guidelines for designing structures with improved mechanical responses.

16. Abstract German

In dieser Arbeit werden die Herstellung, die mikrostrukturelle Entwicklung und das mechanische Verhalten von offenzelligen Aluminiumschäumen und Gitterstrukturen untersucht, die mittels Feingussverfahren hergestellt wurden. Aluminium-Zellmaterialien sind aufgrund ihrer hohen spezifischen Festigkeit und Energieabsorptionsfähigkeit von großem Interesse für Leichtbauanwendungen. Die kombinierten Einflüsse von Gießbedingungen, Legierungszusammensetzung und architektonischem Design auf das Verformungsverhalten solcher Strukturen sind jedoch bislang nur unzureichend verstanden. Durch die Integration der Auswirkungen von Legierungsänderungen in das strukturelle Design liefert diese Arbeit neue Erkenntnisse über die Zusammenhänge zwischen Prozess, Struktur und Eigenschaften von gegossenen Aluminium-Zellmaterialien.

Bei Al-Si-Legierungsschäumen zeigte sich, dass die Formtemperatur einen dominanten Einfluss auf die Erstarrungsmikrostruktur und das Verformungsverhalten ausübt. Höhere Formtemperaturen führten zu dendritischen Mikrostrukturen mit groben eutektischen Siliziumphasen, die an den Stützflächen lokalisiert waren, was zu spröder Verformung, vorzeitigem Versagen und gezackten Spannungs-Dehnungs-Kurven führte. Niedrigere Formtemperaturen hingegen begünstigten globuläre Aluminiumkörner und eine gleichmäßigere Siliziumverteilung, wodurch duktileres Verformungsverhalten und glattere Spannungs-Dehnungs-Verläufe ermöglicht wurden. Diese Vorteile wurden jedoch durch verfahrenstechnische Herausforderungen abgeschwächt, da eine unzureichende Formfüllung und dünnere Streben die Eigenschaftsverbesserungen der unter diesen Bedingungen hergestellten Schäume begrenzten.

Lokale Abkühlratenvariationen während des Schaumgusses beeinflussten zusätzlich feinere mikrostrukturelle Merkmale. Obwohl die Abkühlungsgeschwindigkeit insgesamt nur moderat variierte, verbesserte eine schnellere Abkühlung die Wirksamkeit der Strontiummodifikation, verfeinerte die eutektische Phase und veränderte die Morphologie eisenreicher Intermetallphasen. Diese Ergebnisse verdeutlichen, dass die Empfindlichkeit gegenüber mikrostrukturellen Veränderungen insbesondere steigt, wenn die Erstarrung bei niedriger Abkühlrate und in begrenztem Erstarrungsraum erfolgt, wo lokale Diffusion und Erstarrungsdynamik die Phasenbildung stark beeinflussen.

Das Verformungsverhalten erwies sich zudem als stark abhängig von Größe und Verteilung der Sekundärphasenpartikel innerhalb der mikrometergroßen Streben. Grobe, spröde Phasen wie eutektisches Silizium führten zu Dehnungslokalisierung und vorzeitigem Versagen, während die Verfeinerung dieser Phasen oder deren

Ersatz durch feine Ausscheidungen, wie in wärmebehandelten Al-Cu-Legierungen erzielt, eine duktile und homogene Verformung förderten. In Schäumen mit gleichmäßiger und verfeinerter Mikrostruktur korrelierte die Plateauspannung direkt mit intrinsischen Materialeigenschaften wie der Strebenhärte und zeigte ähnliche Trends wie in Massivlegierungen. Dies zeigt, dass bei Minimierung spröder Bestandteile die mechanische Antwort auf Schaumniveau von fundamentalen Materialparametern bestimmt wird.

Durch den Einsatz indirekter additiver Fertigung zur Erzeugung definierter Zellgeometrien konnte das Zusammenspiel zwischen Architektur und Materialverhalten weiter untersucht werden. Die Ergebnisse zeigten, dass die relative Dichte in Strukturen mit dickeren Streben weniger dominant ist, da hier die Verformung primär vom Zelldesign bestimmt wird. In biegedominierten Gittern konnte sprödes Versagen ohne Biegen auftreten, wenn eine ungünstige Legierungschemie oder Gießfehler vorlagen, was verdeutlicht, dass architektonische Vorteile durch eine schlechte Mikrostrukturqualität aufgehoben werden können. Bei größeren Strebendimensionen hatten feinskalige mikrostrukturelle Modifikationen nur geringen Einfluss auf die Verformung. Mit abnehmender Strebendicke nahm die mikrostrukturelle Empfindlichkeit jedoch deutlich zu, was eine Wechselwirkung zwischen Geometrie und Materialverhalten aufzeigte. Darüber hinaus erwiesen sich zugbelastete Elemente in gegossenen Gitterstrukturen als mechanisch ineffektiv, was darauf hinweist, dass solche Merkmale im Design vermieden werden können, um die Effizienz zu erhöhen.

Insgesamt verdeutlicht diese Arbeit, wie Verarbeitung, Mikrostruktur und Geometrie zusammenwirken, um die Leistungsfähigkeit von gegossenen Aluminiumschäumen und -gittern zu bestimmen. Durch die Verknüpfung von Legierungsdesign und Strukturverhalten über mehrere Längenskalen hinweg wird ein umfassender Rahmen geschaffen, um leichte zellulare Metalle durch kombinierte mikrostrukturelle und geometrische Optimierung gezielt anzupassen. Die Ergebnisse leisten einen Beitrag zum besseren Verständnis gegossener zellulärer Aluminiummaterialien und liefern praxisnahe Richtlinien für die Gestaltung von Strukturen mit verbesserten mechanischen Eigenschaften.

