

École polytechnique de Louvain

Influence of contour parameters on the fatigue properties of AlSi10Mg produced by L-PBF

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Abstract

The presence of porosities and low surface quality of samples manufactured by L-PBF is one of the main challenges faced by the additive manufacturing industry. These surface and sub-surface defects have a detrimental effect on the mechanical performance, especially under cyclical stress. Traditional post-treatments aim to refine the microstructure, improve surface quality, or alter the stress state of the surface in order to improve the fatigue performance. The main drawback of these processes, however, is that they are time-consuming and may not lead to significant improvements in fatigue lifetime. An alternative to these post-treatments is the optimization of the contour process parameters. By increasing the contour track energy density, it is possible to reduce the surface roughness, but at the expense of the creation of keyhole porosities. In this investigation, the influence of roughness-optimized contour parameters on the fatigue behavior of net-shaped, vertically-built AlSi10Mg samples is studied.

Results show that a high track energy density will lead to an improved surface roughness, but the appearance of keyholes will depend on the melting mode. Operating at a TED of 912 J/m (laser power: 272.6 W, scanning speed: 300 mm/s) leads to a keyhole melting mode, and results in a surface roughness of 4.2 μm and keyholes in the contour. In contrast, operating at a TED of 456 J/m (laser power: 272.6 W, scanning speed: 600 mm/s) leads to the onset of a keyhole melting mode, and results in a surface roughness of 5.1 μm and no keyholes in the contour. The lower roughness was attributed to an increased wettability of the lateral surface, and the keyholes to deeper and more unstable melt pools that trap gas bubbles.

An optimization process has shown that, for the same track energy density, the geometry of cylindrical fatigue sample will influence the roughness level and appearance of keyhole porosities. Larger cross sections are associated with a greater total heat input and less heat dissipation, which leads to variations in roughness levels and porosity distribution along the height of the sample.

A stress-relief heat treatment of 250°C for 2 hours will lead to a decrease in the ultimate tensile strength of approximately 450 MPa in as-built samples to 410 MPa in heat treated samples for all contour conditions. The heat treatment caused limited globularization of the Si network. Furthermore, it is likely it affected the contributions of Si precipitates within the Al matrix to the strain hardening capacity. These factors could explain the drop in ultimate tensile stress before and after the heat treatment. No differences in the tensile properties in either AB or SRHT samples could be attributed to the contour parameters. Overall, the heat treatment did not alter the microstructure significantly.

Stress-relieved samples manufactured using a contour TED of 912 J/m (100-300) and 456 J/m (100-600) had a better fatigue performance than the reference samples, manufactured using a TED of 193.8 J/m in the studied stress range. This improvement can be attributed to the lower surface roughness in both sample groups due to the contour processing parameters. The fact that the best-performing samples were the 100-300 (best surface roughness and keyhole porosities) indicates that these porosities do not negatively impact the fatigue performance. Further improvements in fatigue performance were achieved by polishing the samples. However, polishing exposed the keyholes in the contour of the 100-300 samples, causing a lower performance than polished 100-600 samples. Sandblasting produced the best fatigue performance in 100-300 and 100-600 samples due to the introduction of residual stresses into the surface. However, it presented a low repeatability and high scattering of the results.

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1. Introduction

Additive manufacturing (AM) is a term that encompasses the collection of processes in which a physical object is built to shape through a layer-by-layer deposition and consolidation of the constituent material. The novelty behind this approach lies in the use of two components: the first one is a digital model (usually built from a computer-aided design file) which contains all the necessary information about the geometry of the sample. The second one is a digitally controlled, part-independent manufacturing system with adjustable parameters, which reads the information contained in the design file and transforms the base material (in liquid or solid form) into the finished sample. Originally conceived as rapid prototyping methods due to their versatility, AM technologies are now capable of generating fully functional components in a wide range of materials in the transportation, aerospace, pharmaceutical, sports, energy, and construction sectors, among others [1-2].

In the context of components for engineering applications, metallic AM has gained prominence in the past few decades. The two main categories of metallic AM that are currently being used are Direct Energy Deposition (DED) and Powder Bed Fusion (PBF). These two technologies rely on the use of a specially prepared metallic feedstock (usually powder or wire) and a high energy source. Through the gradual deposition, fusion and solidification of successive layers of metallic alloy, the part is built. Unlike traditional manufacturing technologies, AM offers several advantages such as a direct translation of concept to component, development of components with a higher degree of customization and complex geometries, reduction in material waste, use of part-independent tools, the ability to produce parts which require little-to-no additional processing, and a significant reduction in overall development and manufacturing time [3].

Despite this, there are still challenges which need to be addressed in the field of metallic AM if it is to become a fully-developed and reliable alternative to traditional manufacturing methods such as casting, molding, etc. There is currently a relatively limited number of metallic alloys which can be reliably printed. Furthermore, due to the melting and solidification dynamics intrinsic to the manufacturing process, there are seemingly unavoidable metallurgical differences between traditionally- and additively-manufactured metallic components. These include undesirable microstructures, residual stresses, mechanical anisotropy, and surface and sub-surface defects. All these manufacturing defects result in limitations regarding mechanical properties such as strength, fracture toughness, ductility, and fatigue life [3-4].

In recent years, researchers have tried to understand the sources of these limitations, how the process parameters and material interact during the printing process, and why they give rise to particular microstructures and properties. Aluminum-based alloys have been given particular attention due to their applicability in critical, light-weight aerospace components. However, unlike steel, titanium, and nickel-based alloys, some of their physical properties such as adherent oxide film formation, high reflectivity and conductivity, thermal expansion coefficient, and a relatively wide range of solidification temperatures have been deemed unfavorable for AM processing [5]. Thus, it has become clear that the goal should be the optimization of process parameters and post-treatments in order to minimize the defects and improve the mechanical properties of additively-manufactured aluminum-based structures.

At UCLouvain, Aluminum-Silicon based alloys have been the object of numerous investigations. Most recent research has been focused on understanding crack nucleation and growth in net-shaped and processed samples [6] as well as enhancing the fatigue properties of AlSi₁₀Mg components produced by laser powder bed fusion (L-PBF) by friction stir processing [7] and hot isostatic pressing [8]. However, the need for post-processing negates one of the main advantages of AM, namely the ability to produce ready-to-use components that, in theory, require no or very little subsequent treatments to achieve an acceptable mechanical performance.

To combat the nucleation of surface cracks that can result detrimental in fatigue life, a traditional way to improve the net-shaped quality of additively-manufactured samples is to use a contour scan on the edges of the geometry. However, comprehensive experimental fatigue testing on specimens manufactured in different orientations and subjected to different surface treatments revealed that the addition of a contour scan is not enough to produce a significant improvement on surface roughness or fatigue life when compared to net shaped samples without contour when coupled with vibratory polishing [9].

To address this problematic, an interesting line of research lies in exploring the simultaneous use of a set of optimum process parameters and scanning strategies which can simultaneously minimize the impact of surface roughness and voids. Recent investigations has shown that a contour-first strategy combined with a level of Track Energy Density (TED) at the onset of keyholing (subsurface defects associated with high energy inputs) provides the best surface finish [10], establishing a promising starting point for future fatigue investigations. Indeed, experimental results focusing on repeated bending of rectangular plates using a contour-first scanning strategy and roughness-optimized contour parameters show a greater fatigue life than density-optimized contour parameters at the same stress level [11]. However, questions still remain as to the impact of the keyhole porosities that are formed, and how these, coupled with the roughness, will impact the fatigue behavior.

Using the previous results as a starting point, the present investigation aims to study the effect of the contour parameters on the uniaxial fatigue life of net-shaped cylindrical samples of AlSi₁₀Mg manufactured by L-PBF. To achieve this, standard fatigue specimens with different central diameters are printed using a variety of roughness-optimized contour parameters at the onset of keyhole and in keyhole conditions. Through optical microscopy and roughness measurements, the optimum combination of dimensions and roughness that ensures keyhole conditions is determined. Mechanical testing, both static and dynamic, is performed on the samples, in order to determine the mechanical properties and construct a fatigue life curve for each of the studied conditions. Further fatigue results on polished and sandblasted samples with the same contour parameters aim to highlight the effect of the surface quality and residual compressive stresses on fatigue behavior. A presentation of the fatigue results is followed by a discussion on how the contour parameters affect the fatigue behavior.

2. State of the Art

2.1. **Laser Powder Bed Fusion (L-PBF)**

Laser Powder Bed Fusion (L-PBF), as a concept, can be synthesized as the layer-by-layer selective fusion of feedstock powder using a high energy source. However, in practice, this technology requires the interaction between multiple material and process inputs that need to work together in order to produce a high-quality, fully-functional structure that meets the design criteria and operational requirements of any given application. Process parameter optimization can lead to an improvement in the mechanical properties and mitigation of detrimental effects caused by the inherent defects of L-PBF. To understand how, it is necessary to review the process, inputs, and their interactions in more detail. The following sections offer an overview of the L-PBF process, the materials currently in commercial use, the characteristics of the powder, and the process parameters.

2.1.1. From a Digital File to a Solidified Structure

The manufacturing process starts with a CAD (Computer-Aided Design) file, which contains all the necessary geometrical information of the sample or structured to be printed. This CAD file is then converted to an STL (Standard Triangle Language) file, in which all the surfaces are described via triangulation. This file conversion is necessary in order to proceed with the slicing process, in which the geometry is first checked for defects associated to errors during the triangulation of the surface, and subsequently sliced into layers with a predetermined thickness depending on the build orientation. During this stage, it is possible to define support structures, which might become necessary depending on the geometry of the sample and/or removal procedures of the samples from the plate. Finally, the slice files are positioned on a virtual build plate, where it is possible to group different samples with different geometries and process parameters into families and define the process parameters and scanning strategies to be used in each one. At this point, the file is ready to be loaded into the machine computer [2].

Once inside the machine computer, the process is entirely automated though manual control is possible in order to ensure that there are no mistakes in any of the building stages of each layer. The manufacturing itself takes place inside an inert construction chamber. The size and particular characteristics of the chamber may vary from manufacturer to manufacturer, but they all share some essential components: an inert gas feed, a powder dispenser, a feedstock powder piston, a construction plate, a build platform piston, a laser heat source, and a scanning system. Before the printing starts, the chamber is first purged of oxygen by setting an acceptable limit value and an inert gas (in most cases, argon), called shielding gas, is introduced. The purge and introduction of a shielding gas is necessary in order to prevent any gas from becoming entrapped inside the powder particles during the powder atomization process and from contaminating the melt pool of liquid metal or the hot solidified structure, which could lead to porosities.

Once ready, the feedstock powder will be supplied through a piston, which is elevated until some of the powder is within the action range of the powder dispenser. In many cases, this element corresponds to a roller, which displaces horizontally in a fixed path that will take it over the feedstock powder and build platform. Once the powder is in position, the roller will displace and move some of the powder with it, covering the build plate with a fine layer before returning to its starting position. This step (called layering) is essential, as a well-distributed and uniform powder layer is necessary in order to ensure that the first layers will have a proper adhesion to the build plate. Moreover, uniform layers will hold the parts in position and will compensate for any internal stresses caused by the thermal expansion which could lead to warpage or delamination.

When the layering is complete, the laser will heat up, and by using a system of articulating galvanometer mirrors, the laser spot moves across the powder bed, fusing the metallic powder in a predetermined scanning path. The laser will impinge into each individual metallic powder within the laser spot, and part of the energy will be absorbed while another part will be reflected into other particles. The amount of heat absorbed will depend on the thermos-physical properties of the powder. As a result, the packed powder fuses into a track of solidified metal that is welded to the previous layer. This process is repeated along the entire scanning path, and laser tracks are melted and remelted, with some overlap which welds them to one another. When this process, called sintering, is complete, the base plate will be lowered by its piston, so that the process can start anew after each successive layer [3]. Figure 2.1 represents this process in a simplified schematic:

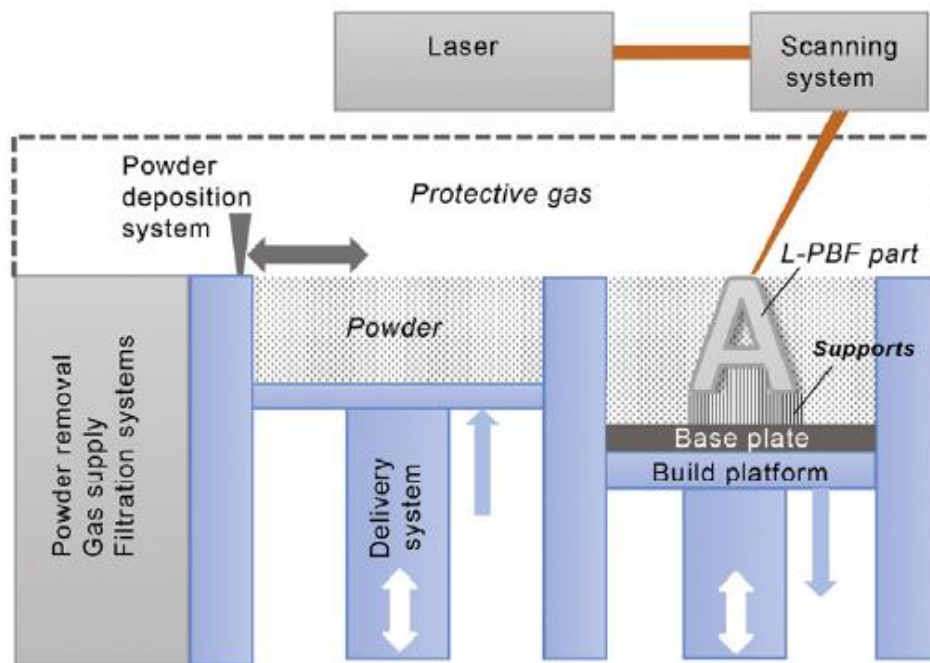


Figure 2.1. Schematic representation of the successive layering and sintering process in L-PBF. [12]

2.1.2. Characteristics of the Feedstock Metallic Powders

The feedstock of metallic powder from which a structure will be built is the fundamental material input in L-PBF. Therefore, the impact of this powder's preparation and characteristics on the part's final quality and properties cannot be underestimated. In traditional manufacturing technologies, it is well-known that the history of the feedstock material and its preparation has an impact on the characteristics of a products manufactured from it. Indeed, after years of both research and experience, nowadays these conventional technologies use pre-fabricated and pre-formed materials that come in standardized shapes and sizes. The thermal and mechanical processing history of these traditional material stocks is well-known and adapted by the supplier to meet the process for which it is intended. Currently, the design of metallic powder feedstock in the AM field is carried out through a combination of property identification, computer-aided multiscale material-manufacturing modelling, and trial and error analysis. As a result, it has become clear that the feedstock preparation needs to be customized to each individual process, and slight variations might result in unforeseen consequences in the solidification and final performance [1].

A better understanding of the metallic powder characteristics comes from taking a multiscale approach and subdividing the powder properties into multiple levels, going from the powder properties of individual particles to the bulk properties of a collection of particles. On an individual scale, powder properties such as morphology, particle size distribution, impurities, composition, moisture, and particle density play the biggest role. At the same time, these properties will be reflected at the bulk scale, where properties such as flowability and powder packing are used to describe the behavior of the whole under the process conditions [13]. Here, emphasis will be placed on the relationships among three key physical properties, namely: morphology, particle size distribution, and flowability.

- **Flowability.** At the bulk level, flowability is probably the most important characteristic of a metallic powder. This term refers to the behavior of the powder when it is set into motion or subjected to stress and is crucial because it ensures the smooth spreading of even powder layers across the entire surface of the build plate, resulting in a consistent thickness and uniform distribution and density. If the flow properties of the powder are not suitable and consistent, then there is a risk that particles might clump together, leading to defects in the printed parts. Therefore, in order to ensure that each successive powder has consistent properties and adheres properly to the previous layer, it is necessary to guarantee, as much as possible, good flowability of the powder. To do this, it is necessary to take a look into the morphology and size distribution of the powder particles.
- **Morphology.** The morphology refers to the shape and supplemental geometrical characteristics of the powder. Even though this characteristic will evidently be affected by several factors, perhaps the most important one is the manufacturing process. Figure 2.2 illustrates the different particle morphologies that result from Plasma Electrode Rotating Process (PREP), Rotatory Atomization (RA), Gas Atomization (GA), and Water Atomization (WA). Based on these observations, it is evident that the morphological qualities differ significantly from process to process. Even though it is hard to make definitive statements (as trends may vary with particle size), it has been observed that coarse particles, smoothness and sphericity favor flowability and deposition. When the particle size decreases to the point where particles become fine, adhesive forces might play a role [13]. In addition to this, it has been observed that irregular particles with coarser textures lead to a rougher surface in the finished part. Thus, particles produced by PREP (a, g) will present a better flowability than those produced by GA (d) and WA (e) [3].

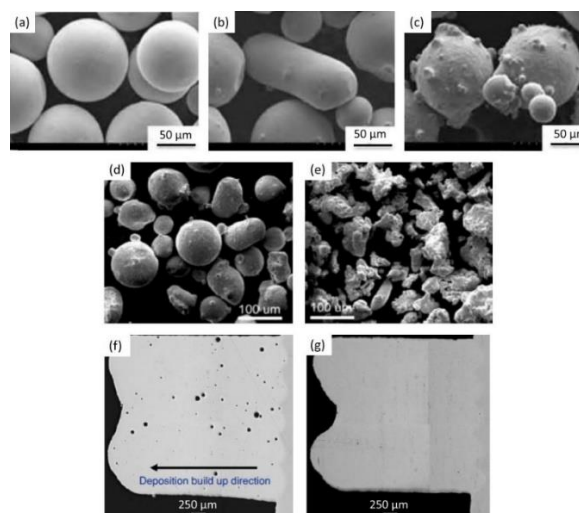


Figure 2.2. Comparison of powder morphology based on manufacturing processes. [3]

- **Particle Size Distribution (PSD).** The PSD quantifies the variations in the size of powder particles. Like morphology, this characteristic is influenced by the manufacturing method, but not as strongly. Research has shown that PSD influences layer density, thermal conductivity, and flowability [13]. In general, finer powders with a more uniform PSD have a better flowability, as they enable a more even deposition and formation of smaller melt pools under the action of the energy source [3].

There are different ways to try to homogenize feedstock powder pre-production in order to achieve the best flowability. The most basic and popular technique is sieving, which consists in separating particles of different sizes using a filter. Based on the available equipment this process can be time consuming. Nevertheless, powder homogenization becomes necessary when dealing with recycled powder. One of the main advantages of the L-PBF process is the ability to reduce material waste, which is accomplished by collecting the powder that is not fused during the sintering and reusing it for future parts. However, the properties of recycled powder are not necessarily the same as those of new powder. The sintering process might cause clumps of melted powder particles to form, which will be greater in size and irregular in shape. This could explain why some recycled powders have a lower flowability than new powders [13].

Another factor to consider is the amount of powder necessary. As seen before, the quality of the powder depends greatly on the manufacturing process. However, processes which produce a higher-quality powder also produce the lowest yields. Therefore, powder considerations, from manufacturing to conditioning, need to account for the amount of powder needed, the required quality of the final part, the available amount of virgin powder, and the time constraints. Oftentimes, this will lead to a compromise, in which the final selection will depend on the application and available resources.

2.1.3. Metallic Alloys used in L-PBF

For any given application, the material selection process is based not only on the purpose the final part is meant to serve, but also on the mechanical characteristics and processability of the base material. This is especially true for feedstock powders, as their composition will determine the way the L-PBF process parameters are set and the powder is prepared. The role that physical characteristics of the metallic powders has been discussed in Sections 2.1.2., and the role of process parameters will be discussed in Section 2.1.4. It is important to highlight the interactions between these two factors aspect and the material's thermal characteristics. These interactions will control the dynamics of fusion and solidification in the melt pool, which will in turn have an impact on the microstructure, mechanical performance, and presence of defects. Even though the number of alloys currently in service for AM purposes is limited, those who have been deemed as suitable for welding have been found to be suitable for L-PBF. These alloys, subjects of multiple investigations, have found widespread adoption in different fields. Here, a brief overview of the most common alloys used in L-PBF is presented, and AlSi₁₀Mg, the subject of this investigation, is discussed more in-depth.

Steel Alloys. Steels and other iron-based alloys are the most commonly used materials for structural applications due to their high strength and toughness. Among the different steel families, low carbon stainless steels have become especially attractive due to their low cost, ease of processing, and corrosion resistance in extreme environments. Exploring the use of these elements as feedstock material for AM is relevant due to the inherent mechanical limitations induced by the low cooling rates of traditional manufacturing methods such as casting. Furthermore, it allows for the manufacturing of more complex components, such as lattice structures. Unlike the case of

other alloys, research has shown that AM is able to enhance its strength without reducing the ductility of these alloys. Several investigations point out that both the yield and ultimate strength of steel components is increased, due to a more refined dendritic microstructure that provides a higher resistance to dislocation motion, product of faster cooling rates during the process. However, some issues still persist in the AM of steel alloys. From a mechanical perspective, the fracture toughness and fatigue life of these alloys is inferior to that displayed by those manufactured using traditional methods. In addition to this, factors such as the carbon content, presence of low melting alloying elements, and poor flowability of the powder will induce issues such as cracking susceptibility, higher residual stresses, and higher porosity contents [14].

Nickel Alloys. Nickel-based superalloys are primarily used in the aerospace industry. They possess a combination of high-temperature strength, high toughness, creep, oxidation, and corrosion resistance. These qualities make them ideal candidates for use in critical applications under extreme environmental conditions. There has been an increasing interest in using AM to manufacture aerospace components from these alloys, as these tend to have complex geometries, which have been proven to be difficult and expensive to machine without finely-tuned parameters and increasingly specific equipment. Therefore, several investigations have been conducted with the aim of characterizing the influence of process parameters on their mechanical properties. Their results show that aspects such as the energy input, scanning strategy, and build direction will have the biggest an impact on the static and dynamic performance of the alloys. However, it is evident that these properties are still well below those of traditionally-manufactured alloys. The reason for this is the high sensitivity of Nickel alloys to phase precipitates, texture, and grain size, which might lead to non-desirable microstructures that limit mechanical performance. Therefore, in order to make additively-manufactured nickel alloys more competitive, additional post-processing, in the form of heat treatments and other processing techniques, is required [15].

Titanium Alloys. Titanium alloys, particularly β -type alloys, are widely used in the biomedical industry as base material for implants. Unlike stainless steel, which is susceptible to cracking, and cobalt-chromium alloys, which are potentially carcinogenic, titanium alloys have excellent biocompatibility coupled with a low modulus, high specific strength, and good corrosion resistance. The manufacturing of biomedical components by means of traditional manufacturing methods has always proved challenging, mainly due to the difficulty in melting, oxidation and processing. Therefore, the use of AM as an alternative way to produce cellular-type implants with comparable mechanical properties in a faster and cheaper way has been a subject of research in the past few decades. It has been shown that the main problematic lies in the formation of undesirable defects, particularly porosities. When these can be limited by ensuring a sufficient energy input, titanium alloys produced by L-PBF show a comparable and in some cases better mechanical performance to those manufactured using traditional methods. Some issues that remain unclear are the influence of the manufacturing method on the corrosion resistance, and the need for further post-processing to enhance biocompatibility [16].

Aluminum Alloys. Aluminum is the second most-used structural material behind steel. Owing to its high strength-to-density ratio, it has gained widespread adoption in sectors that require lightweight structural components such as aerospace and transportation. In contrast to the other alloys discussed so far, aluminum alloys have traditionally been optimal for low-cost production using traditional manufacturing technologies. Furthermore, some of their inherent properties, such as their proclivity to form surface oxide films, high thermal conductivity, low absorption, and wide range of solidification temperatures make them challenging to process by AM. Despite this, alloying aluminum with silicon, together with minor quantities of magnesium and copper, can

greatly improve the printability of these eutectic alloys. Several investigations have focused on the properties of the Aluminum-Silicon alloys produced by L-PBF. It has been shown that tensile properties such as ductility and strength of Al-Si alloys are better compared to their cast counterparts, owing to the morphology of the Si phase in the eutectic regions. In contrast, the fatigue properties of these alloys are below their cast counterparts, mainly due to the presence of residual stresses, porosities, and unfused particles. Additionally, chemical heterogeneities have been observed as a result of the selective vaporization of alloying elements [17].

Among the different alloys in the Aluminum-Silicon family, $\text{AlSi}_{10}\text{Mg}$, a hypoeutectic alloy, has been the focus of numerous investigations. When processed by L-PBF, this alloy presents a much finer eutectic microstructure compared to that observed using traditional manufacturing methods. The melt pools are composed of two types of grains: elongated columnar grains along the build direction that have a fiber texture along the scanning direction in and near the center of the melt pool, and smaller equiaxed grains with random orientations near the boundaries of the melt pool. Inside the grains, cellular dendritic Al-Si networks are present and can be observed in Figure 2.3, whose morphology varies from fibrous networks close to the center of the melt pool to discontinuous Si particles in the heat-affected zone [18]. Due to the crystallographic nature of the melt pools induced by the rapid cooling and solidification dynamics induced during the L-PBF process, it is important to look into the resulting mechanical properties of built components.

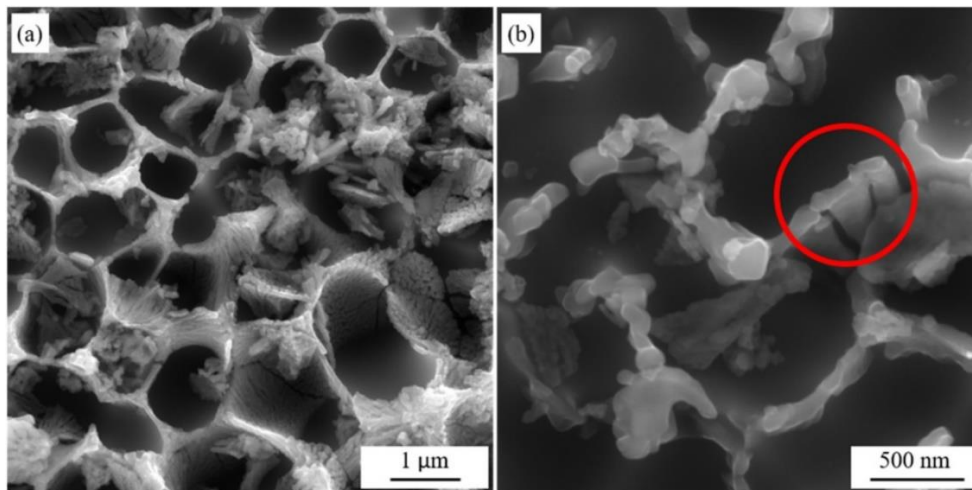


Figure 2.3. a) Network and b) fibrous morphology of the Si phase. [18]

According to research conducted in the past decade, the static properties of $\text{AlSi}_{10}\text{Mg}$ are comparable and in some cases better than those obtained using traditional methods such as casting. As expected, the textured crystallographic nature results in anisotropic properties in the longitudinal (vertical) and transverse directions. However, in both cases, the ultimate tensile strength is higher than that of the cast material, and the elongation at break is comparable. Researchers believe that this is the case due to the fine microstructure of the alloy, coupled with the precipitation hardening of a Mg_2Si phase [19]. Additional research has shown that this alloy experiences gradual yielding, and that the build orientation will have an impact on both the elongation at break and the ultimate stress, which in some cases is higher than that of wrought aluminum alloys, as illustrated in Figure 2.4a [20].

Despite their good static properties, one major mechanical limitation of $\text{AlSi}_{10}\text{Mg}$ lies in its fatigue properties, as illustrated in Figure 2.4b. One researcher found that, compared to wrought Al6061, the fatigue endurance of $\text{AlSi}_{10}\text{Mg}$ was approximately one order of magnitude lower for a

comparable stress level. They attributed this lower fatigue resistance to the presence of large surface voids, which proved to be sites of crack nucleation [20]. These results have been expanded upon by subsequent research, which found that surface and sub-surface defects have a first-order impact in fatigue resistance, as they become sites for crack nucleation [21].

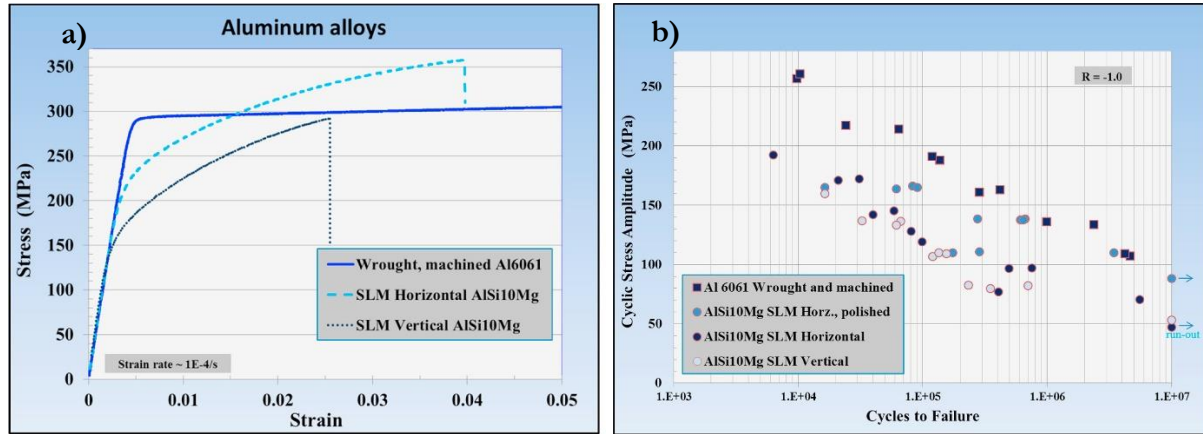


Figure 2.4. a) Stress-strain and b) Wöhler curve of AlSi10Mg and Wrought Al6061. [20]

2.1.4. The Influence of Process Parameters in L-PBF

The fundamental building unit of the L-PBF process is the melt track: the combination of multiple tracks forms a layer, and numerous layers stacked and welded to each other form the final object. Thus, for a given metallic feedstock with a certain set of thermal properties, the way that the energy is transferred from the laser to the powder will determine the laser-material interaction that takes place during melting and the subsequent cooling. This interaction will in turn impact the mechanical properties, microstructure, and defects in the finished part. The most important parameters are the following:

1. **Laser Power.** The laser power refers to the total amount of energy emitted by the laser in a period of time. Most modern L-PBF systems use a continuous wave Ytterbium-doped fiber laser, whose power output can be controlled by an electrical current, and whose beam diameter exhibits a Gaussian distribution [12].
2. **Scanning Speed.** The scanning speed refers to the speed at which the focal spot is moved across the powder bed along a scan vector, usually defined by point distance and exposure time. In most modern L-PBF systems, the deflection of the beam is controlled by two F-theta lenses, which make the position of the focal spot depend in the focal length and deflection angle [12].
3. **Hatch Distance.** The hatch distance illustrated in Figure 2.5e, refers to the spacing between neighboring scan vectors, which should account for a certain degree of re-melting of the previous weld track to ensure full coverage of the region to be sintered [22].
4. **Offset.** The offset refers to the distance between the STL geometry and the scanning vector paths or to the distance between the bulk and contour scanning vectors.
5. **Layer Thickness.** The layer thickness refers to the vertical distance that the collection of welded tracks encompass. It is defined by the vertical displacement of the build plate piston after the end of each layering and sintering cycle [22].
6. **Scanning Strategy.** The scanning strategy refers to the scanning vector geometry and sequence that the focal spot will follow to melt the powder and is aimed at homogenizing the temperature gradient across the entire surface and reducing internal stresses. It encompasses the hatching distance and offset, as well as the angle at which tracks in

adjacent layers are oriented and the order in which the scans are carried out. In the context of this investigation, the scanning strategy will refer to the use of a pre- or post-contour scan. A contour scan can be understood as a single scanning trajectory that encompasses the outer edges of a layer and is distinct from the infill or bulk scan that encompasses the central regions of the layer. From the laser setup, it is possible to define a contour scan to be performed before or after the bulk scan. These are referred to as contour-first (CF) and bulk-first (BF) scanning strategies, respectively. Figure 2.5 illustrates the scanning strategies most commonly used in the industry.

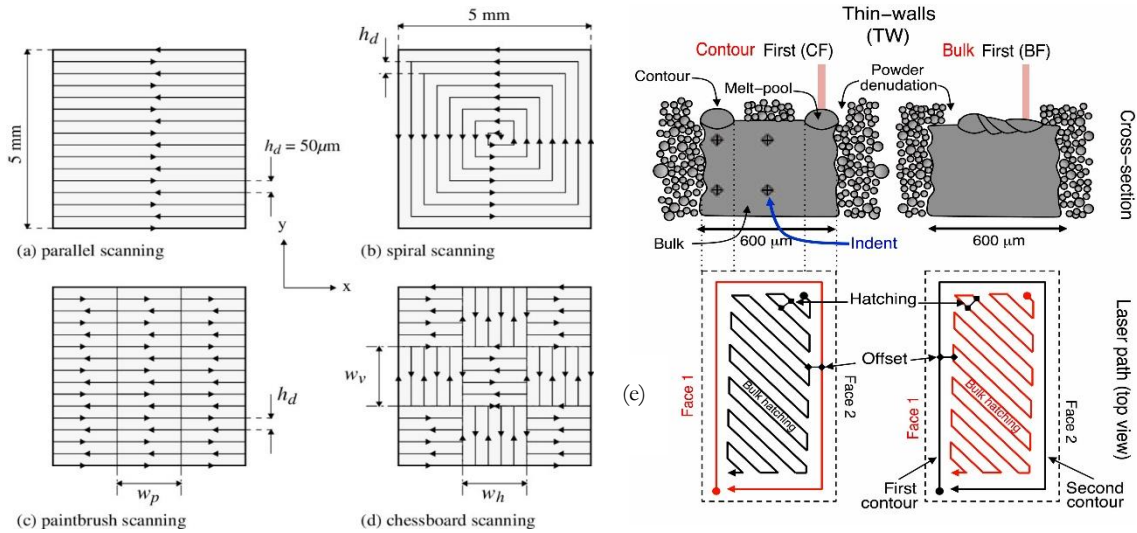


Figure 2.5. a) Parallel, b) spiral, c) paintbrush, d) chessboard, and e) BF, CF scanning strategies. [10] [23]

All the process parameters listed above are independent from each other, which means that there are many possible combinations to achieve the desired operating conditions. Regardless of how these are achieved, the objective is the same: operate in conditions which will ensure that the thermal gradient and solidification structure that arises from their interaction with the metallic powder gives as a result parts with the best quality. Therefore, it is essential to understand how these process parameters affect a key aspect in L-PBF, the melting mode, which will have an impact on the melt pool characteristics and on the defects that might be generated.

The melting mode essentially refers to the way that heat is transferred from the laser to the material during the fusion process. It is possible to establish an analogy between L-PBF and welding, as these two manufacturing technologies share the same common basis: a high energy input is used to induce a phase transformation in the metal that will cause it to become fused to the neighboring material. Thus, although heat transfer is a complex process that depends on numerous factors, not all of them related to the process parameters, it is possible to classify heat transfer in three main categories: conduction mode, transition mode, and keyhole mode. Since the transition mode is simply a term used to describe an operating window in which the heat transfer mode is a combination of the two other modes, it's these who are the subject of most research.

- **Conduction Mode.** In conduction mode, the energy density input is relatively low, and thus the dominant mode of heat transfer is through heat conduction. Some material vaporization close to the surface might take place, but it is usually negligible and does not affect the overall heat transfer process. This melting mode is characterized by little turbulence, little to no spattering, and semicircular melt pools, whose depth/width ratio is lower than 0.5, as illustrated in Figure 2.6a [24].

- **Keyhole Mode.** In keyhole mode, the energy input is relatively high, and thus the dominant mode of heat transfer is convection. This is because a deep vapor cavity forms within the melt pool due to a highly localized energy input, which leads to the vaporization of the material and/or alloying elements. In this mode, the depth of the melt pool is controlled by the recoil pressure that is generated by this localized vaporization. The top region of the melt pool is wide while the bottom is narrow and resembles a keyhole, as illustrated in Figure 2.6b. The depth/width ratio is greater than 0.5 [24].

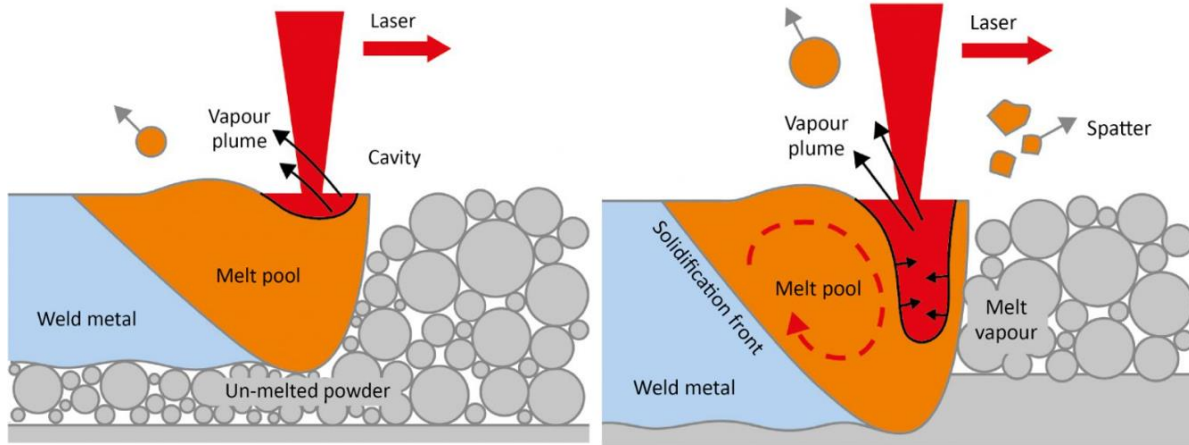


Figure 2.6. a) Conduction and b) keyhole melting mode in L-PBF. [22]

The melting mode is controlled by the energy input, which can be in turn thought of in terms of the process parameters. Even though all of the laser parameters will have an impact on the melting mode, the two which are the focus of the present investigation are laser power and scanning speed. In order to illustrate their effect on the melting mode, consider Figure 2.7, which correlates different non-dimensional laser power-scanning speeds operating windows to their respective melting mode for a variety of aluminum, steel, and titanium-based alloys. From these results, it is possible to state that a high energy input will be associated with a high laser power and/or a low scanning speed, and vice versa, a low energy input will be associated with a low laser power and/or high scanning speed. In order to combine these parameters into a single metric, it is possible to talk about the linear energy input, also known as track energy density (TED), defined as:

$$TED = \frac{\text{Laser Power}}{\text{Scanning Speed}} \text{ [J/m]}$$

Thus, a low TED will be associated with a low energy input and therefore lead to a conduction melting mode, while a high TED will be associated with a high energy input and therefore lead to a keyhole melting mode.

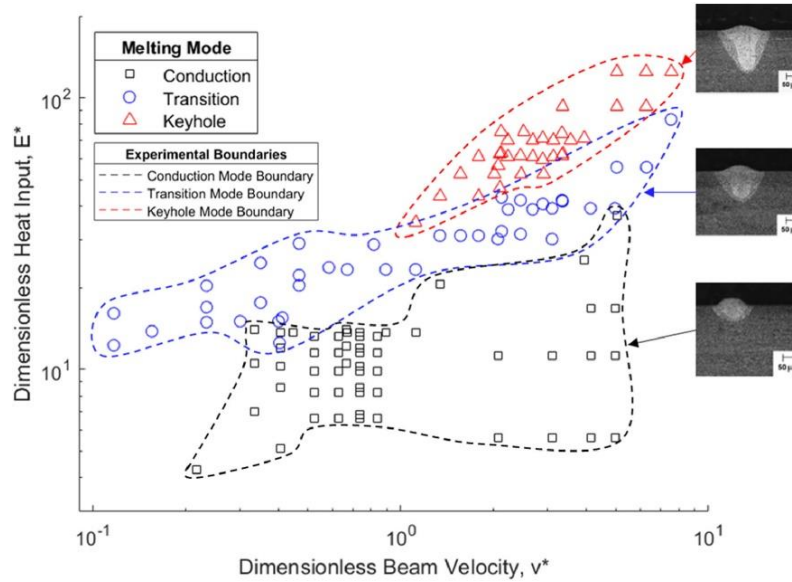


Figure 2.7. Melting modes for different operating windows. [24]

Another important aspect of the processing which determines the way that heat absorbed is the scanning strategy. Simulations as well as experimental results have determined that the choice between a BF or a CF scanning strategy, like those illustrated in Figure 2.5e will have an impact on the melting of powder. In particular, it has been observed that a CF strategy enhances laser absorptivity due to the presence of un-melted powder in the bulk region, which leads to a better melting behavior and continuous track morphology. In contrast, a BF strategy is associated with a lower laser absorptivity coupled with a destabilization of the of the laser-powder interaction, leading to an unstable melt pool [25].

2.2. Surface and Sub-surface Defects in L-PBF

The mechanisms that lead to a material's microstructure, surface quality, and mechanical properties during the L-PBF process are governed by the delicate interactions between process parameters, thermo-physical properties, and powder characteristics. Thus, even slight variations in these factors at any stage have the potential to create unexpected and possibly undesirable effects. For this reason, the generation of defects in parts produced by L-PBF and their effect on the mechanical properties is a trending research topic, since these defects are one of the major factors limiting the mechanical performance, as briefly touched upon in Section 2.1.3. The following sections is a discussion of the two most prevalent defects present in L-PBF: porosities and roughness, how they correlate to the process parameters, and their impact on the mechanical properties of finished parts.

2.2.1. Internal Porosities

The term porosity refers to the ratio between the volume fraction of empty space to the volume fraction of bulk material in a structure. In L-PBF, research has shown that porosities are linked to the process parameters and constitute one of the major factors that limit mechanical performance. This is expected, as they reduce the load-bearing capacity of a structure by reducing the effective area that carries stress, act as sites of stress concentration, and can potentially lead to crack nucleation in fatigue applications. Although numerous pore types have been studied and classified, the two most commonly-occurring types of pores are lack of fusion and keyhole pores, which relate to the melting mode and heat transfer characteristics discussed in Section 2.1.4.

- **Lack of Fusion Porosities.** Lack of fusion porosities occur when the energy input from the laser to the powder bed is insufficient to fully melt the powder layer and remelt the previous powder layers and adjacent tracks, as illustrated in Figure 2.7a. As a result, there is an inadequate welding from one layer to another and from one track to another. The gaps created by the un-melted powder between layers and tracks become pores, which are characterized by an irregular morphology, illustrated in Figure 2.8b. The presence of sharp edges and un-melted powder particles is a clear indicator of a lack of fusion porosity [26].
- **Keyhole Porosities.** Keyhole porosities occur when there is an excessive energy input from the laser into the powder bed, which causes vaporization of the bulk material and/or alloying elements, as illustrated in Figure 2.7b. The vapor depression that forms inside the melt pool is controlled by the recoil pressure, and may become deep and unstable, causing the melt pool to collapse on itself as the laser spot moves away, trapping the vapor inside. They are characterized by a round or quasi-circular morphology as illustrated in Figure 2.8a, and their effect on the mechanical properties will be determined by their number and size [12].

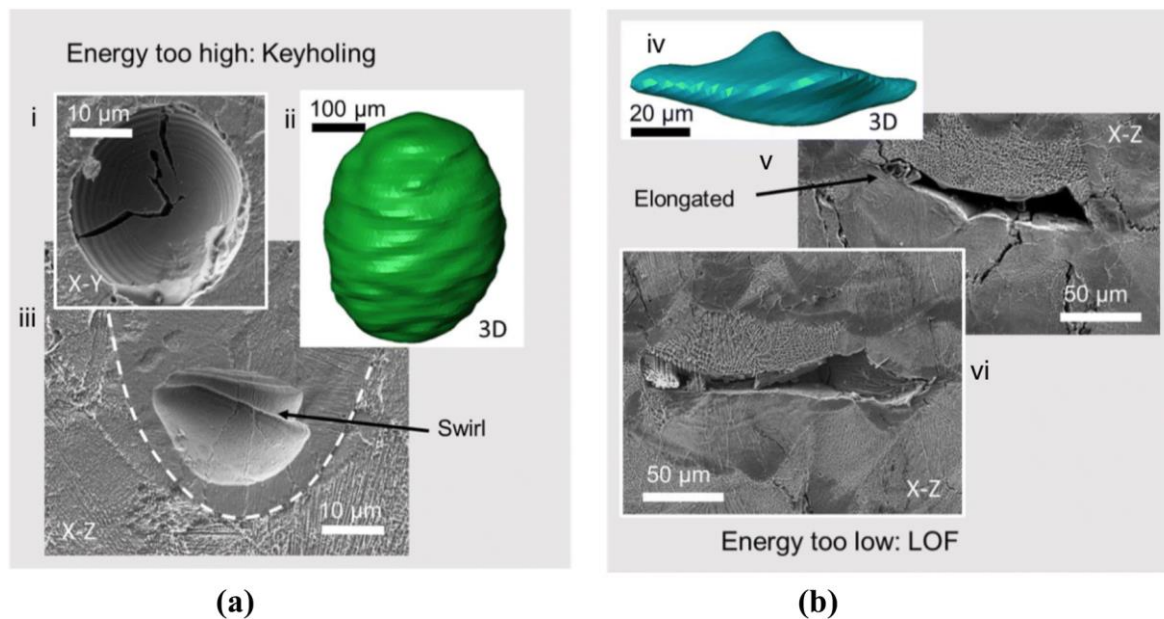


Figure 2.8. Morphology of a) keyhole and b) lack of fusion porosities. [27]

Due to the fact that the formation of porosities is intrinsically linked to the melting mode and its associated dynamics, it is expected that parameters such as the laser power and scanning speed will have the biggest impact on the formation of porosities. However, research into the subject has revealed that the relationship between the melt pool dynamics and these parameters is complex. An investigation into the formation of keyhole porosities and their morphology in titanium alloys illustrates this, as the results show that vapor depressions caused by the vaporization of the elements might be found across a large range of scanning speeds and laser powers as shown in Figure 2.9, where the red and blue lines represent the melt pool and vapor depression transitions, respectively. However, it is important to remark that not all of these operating conditions will result in keyhole porosities. Furthermore, they highlighted the necessity to clearly define operating regimes based on power energy density, as this provided the best way to reliably predict the melt pool stability [28].

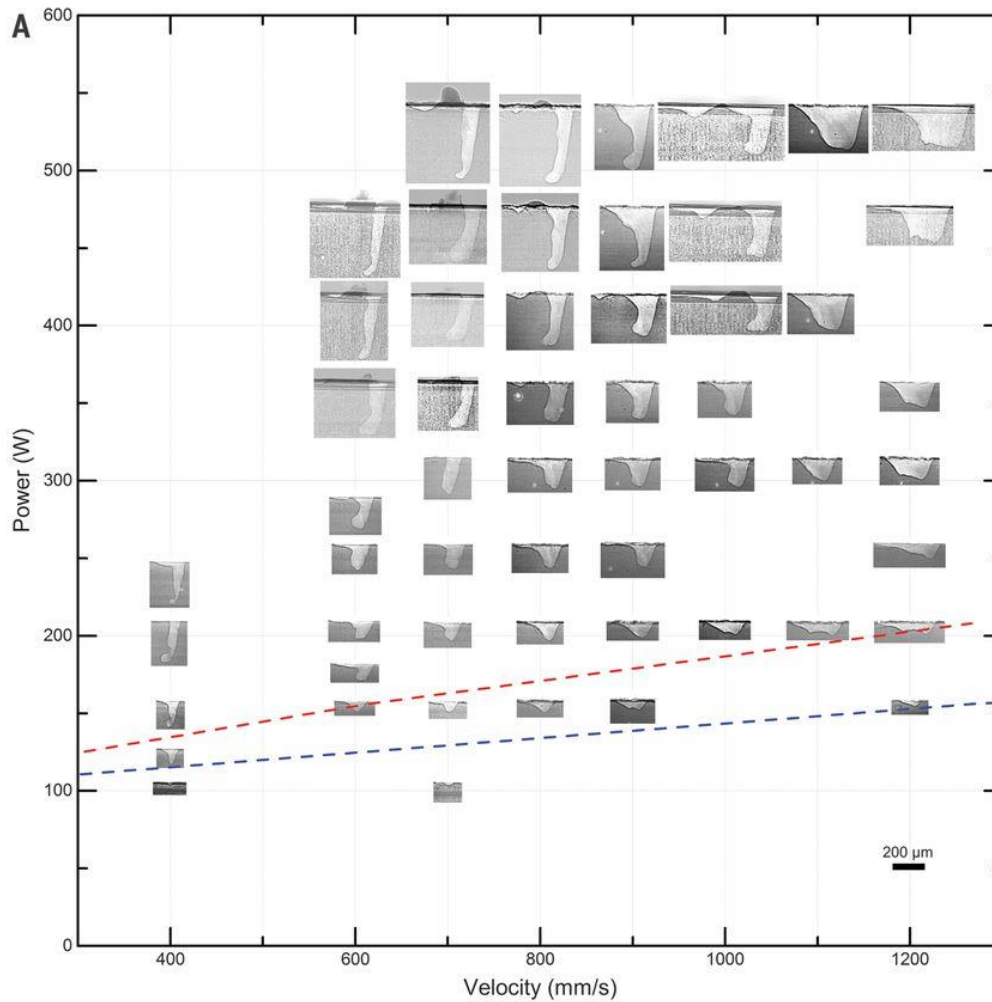


Figure 2.9. Variations in vapor depression size and morphology in the P-V space. [28]

In Al-Si-Mg alloys, researchers have found that the type, orientation and location of defects is associated to different phenomena. For lack of fusion pores, horizontally-aligned defects are caused by an insufficient energy density, which leads to poor fusion with the previous layer. In the case of vertically-aligned defects, an insufficient overlap between scan tracks due to a large hatching distance is believed to be the underlying cause for their appearance. For keyhole porosities, their distribution will depend both on the energy density and scan vector. When using contour and bulk scans, keyholes were observed in the contour as a result of the low heat dissipation into the powder on one side of the tracks, and at the edge of the bulk due to the increased local energy density caused by the deceleration and acceleration of the laser while turning [29].

In recent years, researchers have studied ways to reduce the formation of porosities in finished parts by optimizing process parameters. Experimental results on the effect of varying the scanning speed and scanning strategy on AlSi₁₀Mg show that, for all scanning strategies, scanning each layer twice and pre-sintering the powder to flatten the powder bed produced the most significant reduction in lack of fusion porosities for scanning speeds ranging from 250 – 1000 mm/s [30]. Further studies that focus on optimizing the laser power-scanning speed operation window found that densities above 99.9 % could be achieved by operating at a scanning speed of 1800 mm/s and 400 W, as illustrated in Figure 2.10 [31].

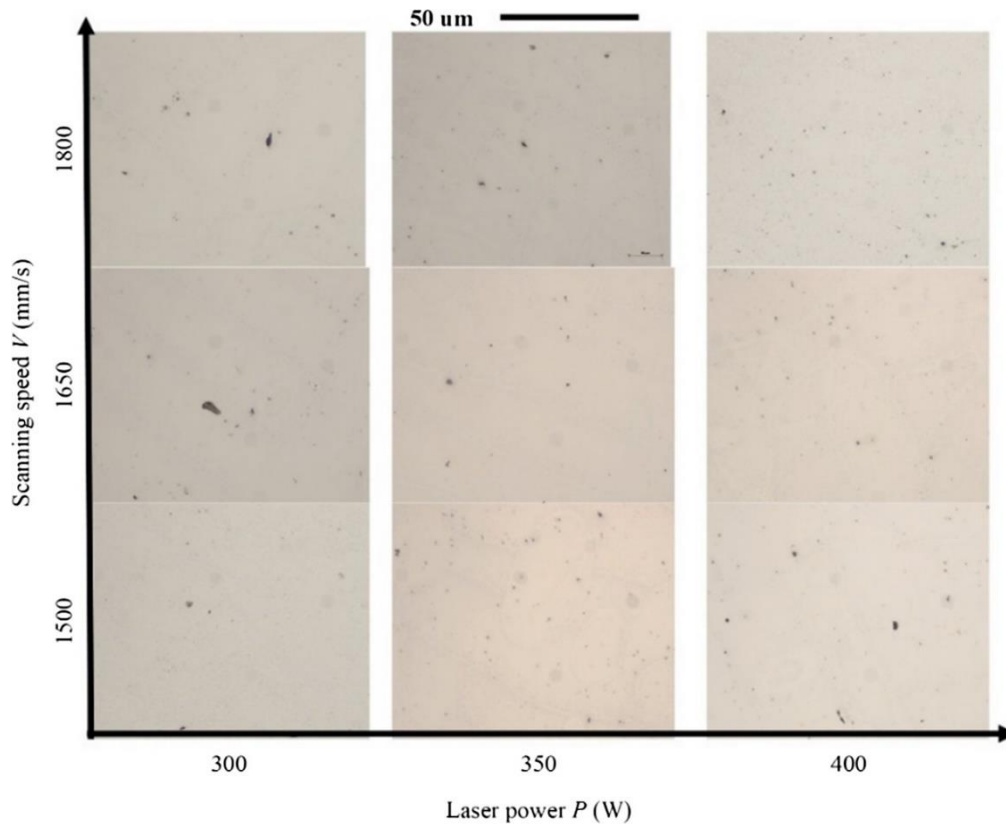


Figure 2.10. Effect of laser power-scanning speed combination on longitudinal porosity [31].

2.2.2. Surface Roughness

Perfectly flat surfaces don't exist in reality, and the nature of the surface can influence the mechanical behavior of a component. However, assessing the quality of a surface is complex due to the multi-scale nature of the deviations that may arise due to the manufacturing process. Thus, deviations are measured at different scales, as illustrated in Figure 2.11. While form and waviness are a measure of the primary and secondary irregularities at a macro- and mesoscale, the most important surface defect in the context of L-PBF applications is surface roughness. Roughness, sometimes referred to as rugosity or asperity, refers to the local measure of microscopic surface irregularities on a part, and quantifies the deviations from the idealized design surface. It constitutes the most significant surface defect in parts produced by L-PBF, and it is common for it to be higher than those of parts produced with conventional manufacturing methods. Investigations have shown that this phenomenon is associated with the thermos-physical interactions between the feedstock metallic powder, part geometry, process parameters, and post-processing. Even though a certain level of surface roughness might be desirable in certain applications, such as the osteointegration of biomedical implants, it is generally detrimental for dynamic load conditions, as fatigue behavior is sensitive to surface quality [12].

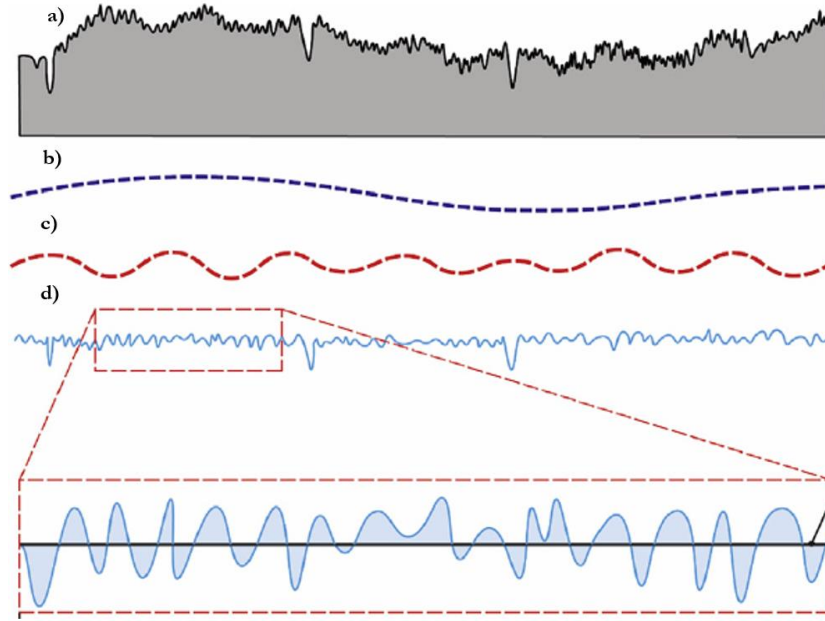


Figure 2.11. Multi-scale nature of surface deviations: a) Texturized surface, b) form, c) waviness, d) roughness. [12]

In L-PBF, the most common way to characterize roughness is by line roughness measurements. To do this, it is necessary to establish a reference, which constitutes the mean line or the equivalent height of the ideal surface, from which the macroscopic deviations are measured, as illustrated in Figure 2.11d. In addition to this, since roughness is a local measurement, a sampling length needs to be defined according to the EN ISO 4288 over which the measurements will be carried out. The most commonly used parameters to characterize roughness are the following [32], and are summarized visually in Figure 2.12:

- i. **Profile Arithmetical Mean Height (R_a):** Indicates the average absolute value of the height of the deviations above the mean line, as illustrated in a. It is calculated as:

$$R_a = \frac{1}{L} \int_0^L |Z(x)| dx$$

- ii. **Profile Peak Height (R_p):** Indicates the point along the sampling length at which the vertical distance above the mean line is highest, as illustrated in Figure 2.12b. It is calculated as:

$$R_p = \text{Max}(Z(x))$$

- iii. **Profile Valley Depth (R_v):** Indicates the point along the sampling depth at which the vertical distance below the mean line is highest, as illustrated in Figure 2.12c. It is calculated as:

$$R_v = \text{Min}(Z(x))$$

- iv. **Maximum Profile Height (R_z):** Indicates the absolute vertical distance between the maximum profile peak height and the maximum profile valley depth, as illustrated in Figure 2.12d. It is calculated as:

$$R_z = R_p + R_v$$

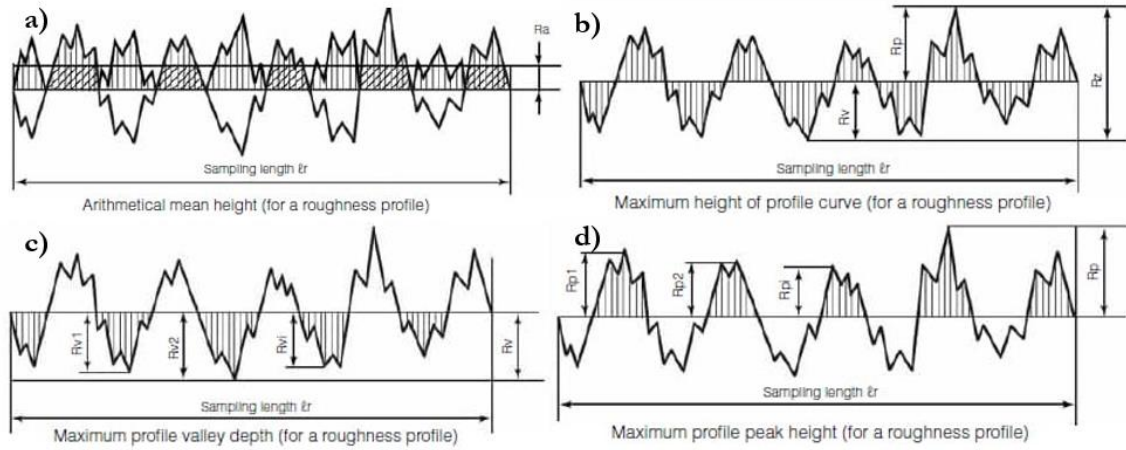


Figure 2.12. Line roughness profile parameters. [32]

Currently, there is no clear consensus on all the factors that affect the surface roughness of a sample produced by L-PBF. This is because of the multifaceted nature of this phenomenon, and the interdependencies which exist among aspects such as surface orientation, track morphology, heat input, melt pool stability and morphology, and powder characteristics. Thus, it is complicated to delineate the boundaries of influence of one aspect with respect to another. In the context of this investigation, the following aspects have been identified as the most influential:

- **The Stair-Step Effect.** The stair-step effect is a limitation of all AM processes that work in a layer-by-layer fashion. It consists in the discretization of a continuous surface into layers, as illustrated in Figure 2.13. This effect is affected by two factors: layer thickness and orientation of the surface. Thin layers result in a lower average roughness but incur longer manufacturing times, such that a compromise must be reached in terms of the tradeoff between time and surface quality [3]. Furthermore, surfaces with low inclination angles with respect to the horizontal have been reported to suffer enhanced stair-step and edge effect, resulting in rougher surfaces compared to surfaces with a higher inclination angle [33].

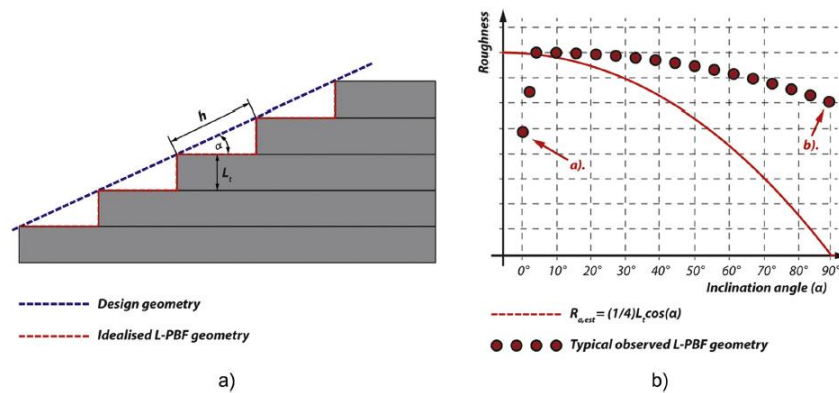


Figure 2.13. a) Illustration of the stair-step effect and b) impact of the surface inclination angle on the roughness. [12]

- **Scanning Strategy.** The use of contour and bulk scans has been shown to have a significant impact on the roughness. By using a contour-first strategy, the laser absorption is increased, thus producing stable melt pools, a continuous track morphology, and better welding from one layer to the previous one, producing a lower roughness. This decreases

the number of locations where overheating occurs, especially for samples with large inclination angles, thus reducing the amount of powder particles that are bonded to the surface [33]. Conversely, a bulk-first strategy produces an irregular and unstable melt pool due to the lack of sufficient melting, leading to the shifting of the melt pool towards the edges and thus promoting the bonding of powder particles (dross) and increasing the roughness, as illustrated in Figure 2.14 [25].

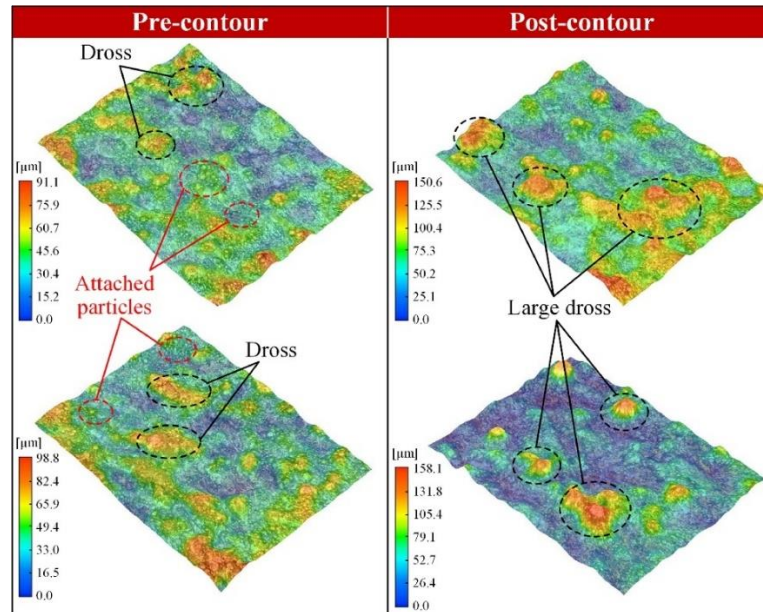


Figure 2.14. Impact of contouring strategy on the bonding of powder particles. [25]

- Energy Input.** The effect of the energy input is related to the melting mode achieved by the process parameters, most notably laser power and scanning speed. If the laser power is low and the scanning speed is high, then the heat input is insufficient to melt the powder particles properly. Under these conditions, the melt pool will become unstable, and will be characterized by an elongation and separation into small balls in order to maintain a uniform pressure inside the melt pool, called balling. As the laser spot advances, these balls may be dragged to the outer edges of the melt pool due to the surface tension gradient of the liquid metal, thus increasing the roughness of at the edges of the tracks, as illustrated in Figure 2.15a [3]. By increasing the heat input, it is possible to stabilize the melt pool, and produce continuous tracks, as evidenced by Figure 2.15c. However, using an excessive energy input can cause an excessive melt pool temperature, resulting in ejected metal vapor which will be compensated for by a recoil pressure, creating ripples on the track morphology, as well as spattering of liquid metal [12].

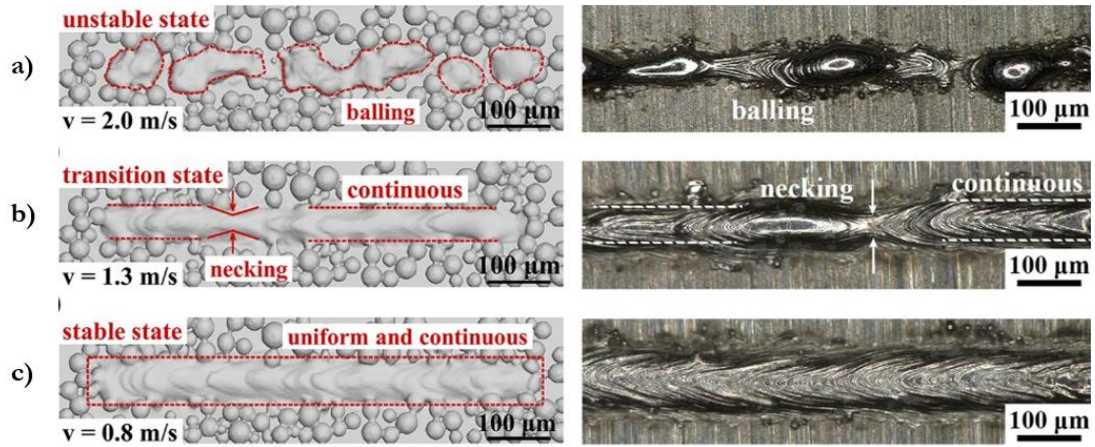


Figure 2.15. Track morphology of a) balling effect, b) transition, c) continuous tracks. [34]

Taking these aspects into account, researchers have tried to find ways to minimize roughness by optimizing the process parameters, as there seems to be a process window corresponding to a high energy input where roughness can be optimized by using a high energy input and an appropriate scanning strategy. By operating in a low TED regime (200-300 J/m), experimental results on AlSi₁₀Mg show that the lateral surface roughness of uniform parts could be decreased by as much as 57 % using a contour-first scanning strategy, with respect to a no contour strategy. Furthermore, researchers found that a bulk -first strategy in this TED regime actually increased roughness by as much as 6 % [35], as evidenced in Figure 2.16.

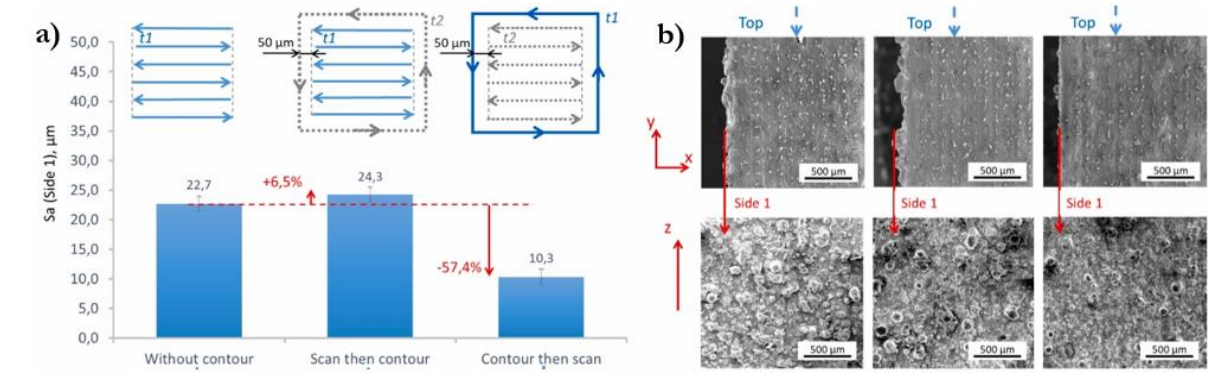


Figure 2.16. a) Roughness levels for different scanning strategies and b) lateral view of the sample morphology. [35]

Despite the improved surface roughness, the R_a values obtained are still significant compared to conventionally manufactured and machined samples. Thus, further investigations have been carried out to determine the optimal window of process parameters that results in the lowest roughness. Experimental results focusing on assessing the impact of hatching space, offset, and TED on thin and ultrathin walls of AlSi₁₀Mg revealed that the optimum roughness levels were achieved at the onset of keyhole conditions. Furthermore, by operating in this regime, using a CF strategy resulted in the TED being the only first-order impact factor on the roughness level. Conversely, using a BF strategy caused the offset and TED to have first-order impacts, and an increased heat input was found to negatively impact roughness. This is illustrated in Figure 2.17 [10].

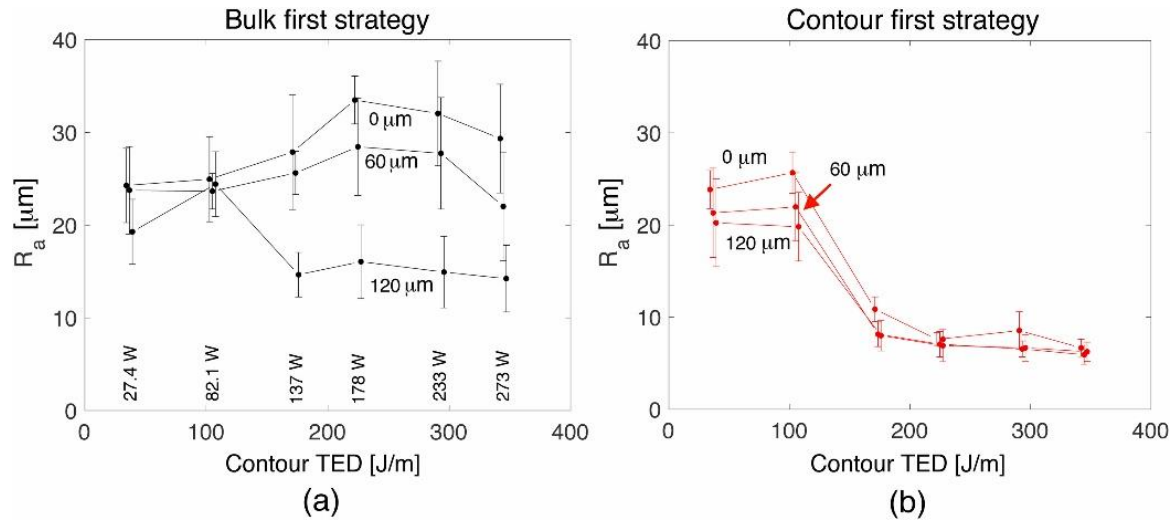


Figure 2.17. Evolution of surface roughness as a function of offset and TED using a) bulk-first and b) contour-first scanning strategies. [10]

2.2.3. Impact of Porosity and Surface Roughness on Mechanical Properties

From the discussion on the different surface and sub-surface defects, it is evident that the current focus of research is finding ways to either reduce or completely eliminate their appearance and prominence within L-PBF parts. This is because they represent inherently weak spots that limit the mechanical performance. Regarding pores, a major concern in current research is the quantification of the impact the different pore types have on the mechanical properties of parts produced by L-PBF. In the case of the tensile properties, results seem to agree that a higher porosity will be correlated with a decrease in elongation, strength, and elastic modulus when porosity levels are above 1%. This is because a large pore concentration will significantly reduce the load bearing area, facilitate void coalescence, act as sites for crack nucleation and stress concentration, and provide a preferential path for crack propagation [26].

However, an important distinction is observed between the effect of lack of fusion pores and keyhole pores. Research has shown that at a comparable porosity level, the reduction in strength and stiffness from lack of fusion pores is much more significant than that from keyhole pores. This is because, as lack of fusion defects occur between layers and have sharp edges, a tensile load would have the effect of opening the crack tips while increasing the stress concentration [36]. Furthermore, a better ductility was observed in parts with a smaller hatching space which experienced remelting and elimination of such pores.

Characterizing the effect of porosity in the fatigue behavior is complicated by the fact that it is greatly affected by several other factors, most notably surface roughness. Nevertheless, investigations generally agree that low-cycle fatigue will be predominantly dominated by surface defects, microstructure, and strength, while high-cycle fatigue will be predominantly affected by the porosity [26]. As with the case of static properties, fatigue properties depend on the size and morphology of the defects. Investigations carried out to determine the effect of the size and type of defects on the fatigue properties of AlSi₁₀Mg reveal that in almost all cases, crack nucleation occurred in lack of fusion defects close to the surface, as evidenced in Figure 2.18, and a correlation was found between the size of the defect and the probability of crack initiation.

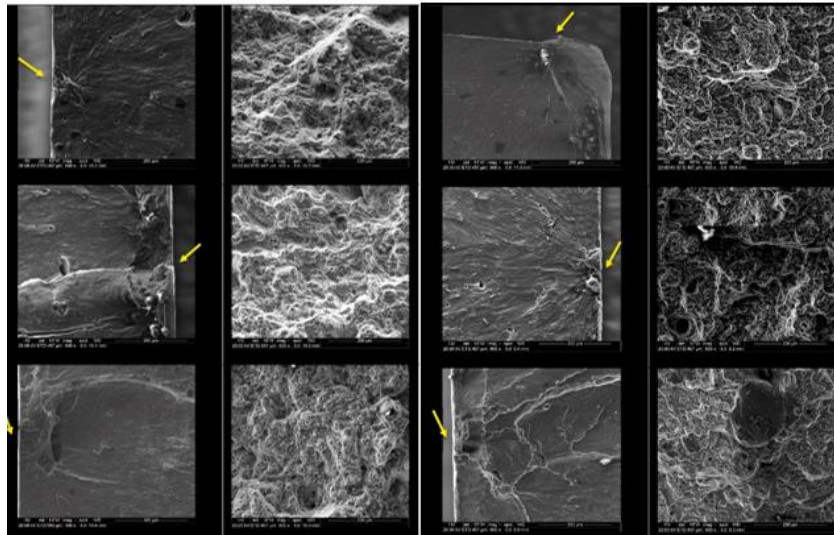


Figure 2.18. Fractography of crack nucleation site and crack propagation surface. [37]

Conversely, well-distributed defects with high sphericity were found to be less detrimental. As in the case of the static performance, the irregular shape of this defect type was found to be an ideal site for stress concentration and crack initiation [6, 37]. These results are backed up by the effect of the powder size, since an increase in the average diameter of the powder particles was correlated to a significantly higher presence of non-melted power, larger lack of fusion pores, and lower fatigue resistance. Furthermore, fatigue crack tips in the same order of magnitude as the space between powder particles were observed [38]. Despite this, some results seem to suggest that sub-surface pores with high sphericity may be more detrimental than initially thought. An investigation into the pore formation mechanisms on $\text{AlSi}_{10}\text{Mg}$ suggests that when a contour scan is used, large round pores at the edge of the bulk due to a localized increase in energy density while the laser is turning could be detrimental to fatigue life [29].

Mitigation of the harmful effect of porosities on the mechanical properties is still a topic under investigation, and researchers are currently investigating the way the morphology and number of pores evolves when heat and surface post-treatments are applied. One investigation focused on the potential benefits of hot isostatic pressing (HIP), a process in which a part is subjected to an elevated temperature and isostatic gas pressure inside a pressure vessel in order to induce microstructural evolution that leads to porosity reduction. Results on the application of HIP with different temperature-pressure combinations on $\text{AlSi}_{10}\text{Mg}$ reveal that this post-treatment, coupled with a T6 treatment was not able to achieve a significant reduction in the volume, size, shape, or number of pores [39]. Further investigations revealed that a minimum temperature of 300 °C in combination with a pressure of at least 150 MPa is necessary to achieve pore closure. However, as illustrated in Figure 2.19a, this leads to an underperformance in fatigue with respect to as-built conditions.

Another post-processing technology, friction stir processing (FSP), has also been proposed as a way to mitigate porosity formation. FSP is a solid-state processing technique in which a rotating tool is plunged into a material, and though the advance of the tool, is capable of forming tracks along which localized microstructural refinement, densification, and homogenization are possible based on the processing parameters [40]. Results at different tool advance speeds on AlSi_{10}M show that this technology is able to achieve a significant increase in fatigue life, as evidenced in Figure 2.19b. The reason for this is the plastic deformation that takes place in the stir zone and leads to the reduction of the of the larger porosities where crack nucleation usually occurs [6].

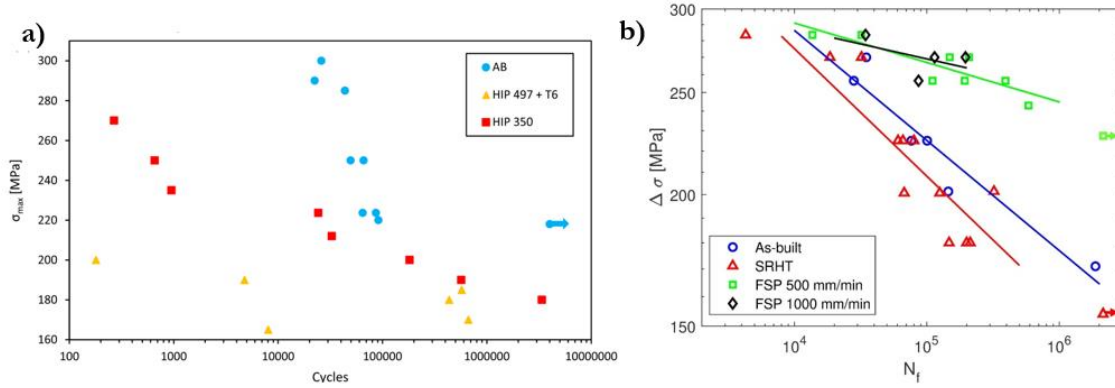


Figure 2.19. Comparison of as-built fatigue behavior of AlSi10Mg with a) HIP [8] and b) FSP [6] post-treatments.

The continuous research on ways to mitigate the detrimental effects of pores are conducted in parallel with those aimed at understanding and reducing the negative effect of roughness that comes from the impact of surface quality on fatigue life. In metals, surface roughness disproportionately affects fatigue life compared to the tensile properties, as the former is dependent on the surface quality, because this is the site where fatigue cracks are most likely to nucleate. Crack nucleation occurs due to the inherent microscopic inhomogeneities at the surface, which can lead to stress concentration and the formation of persistent slip bands. Through the fatigue lifetime of a component, an intrusion-extrusion mechanism will take place along the slip bands, causing the microcracks to grow and start propagating into the material.

Surface roughness is considered to be among the primary factors affecting the fatigue of L-PBF components. It has been generally observed for a variety of materials and microstructures that a decrease in surface roughness, measured in terms of R_a , will lead to an increase of the fatigue limit at all stress levels, though this effect is more significant for specimens with smaller internal defects and coarser lamellar microstructure, and results will present a higher degree of scatter as these internal defects become more prominent in fatigue life at low stress levels [41]. Typically, the way to assess the impact of surface roughness on fatigue life is by assessing the impact of factors that affect surface roughness on fatigue life. Research carried out into the different materials reveals that factors that influence the surface roughness will invariably lead to different fatigue performance behaviors.

One approach to assess the impact of roughness on fatigue life focuses on the impact of the building direction. It is widely known that anisotropy in components produced by L-PBF occurs due to the microstructure that emerges as a result of the layer-by-layer manufacturing. Furthermore, the building direction the major contributor to the staircase effect and its influence on fatigue life. Experimental results on Ti-6Al-4V reveal that inclination angles in the $0 - 30^\circ$ range lead to lower fatigue lifetimes, as they produce downward-facing surfaces. Indeed, the surface roughness of downward facing surfaces was found to be more detrimental than that of upward-facing surfaces, as reductions of up to 35 % in fatigue lifetime were observed for the same stress level when rough downward-facing surfaces were present [42].

Another investigation that looked into the strain-controlled fatigue performance of Stainless Steel 316L revealed that samples built with an orientation of 90° had a roughness approximately 4 times lower than those at 45° , and performed better in strain-controlled fatigue, but worse in stress-controlled fatigue. As illustrated in Figure 2.20. This difference was attributed to the refined microstructure in the building direction of the 90° samples and the superior yield strength of the 45° samples [43].

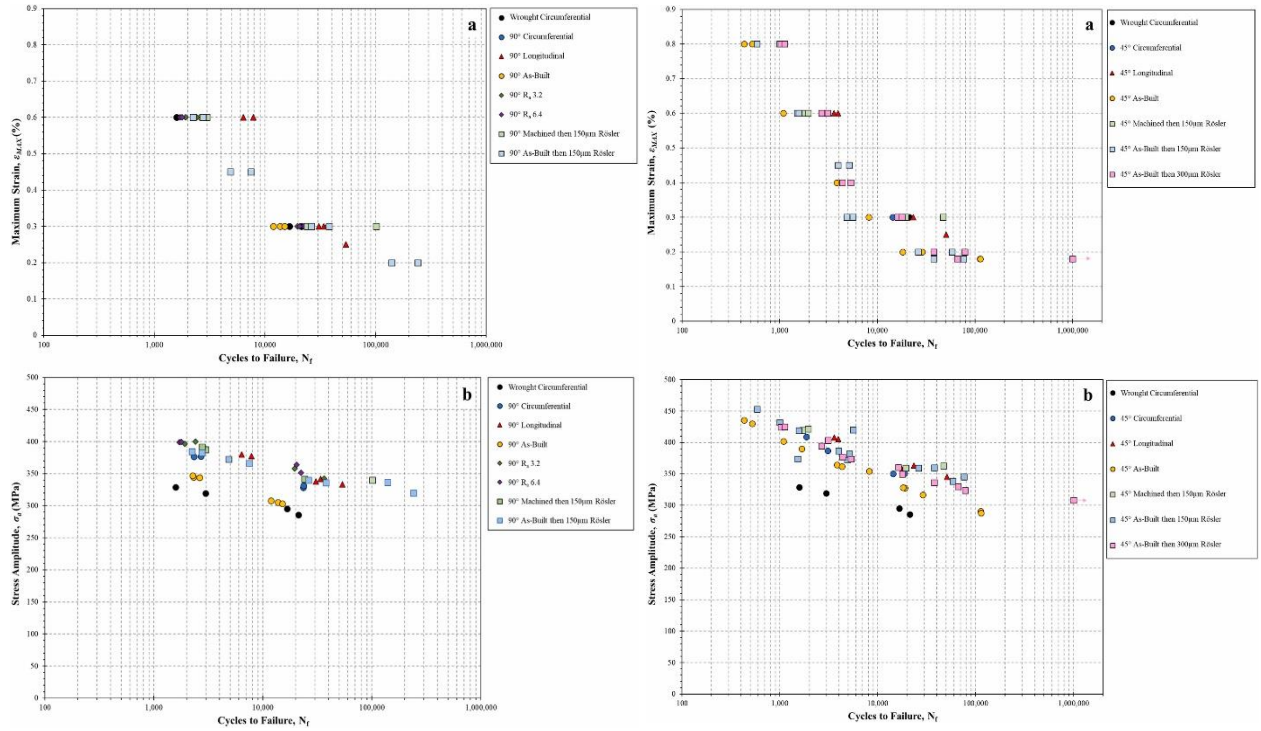


Figure 2.20. Strain and stress-controlled fatigue behavior of samples built at 90° and 45°.

Another innovative approach to increase fatigue life by decreasing the surface roughness consists in combining a contour scan and optimizing the contour parameters. An investigation using A20X revealed that changing the contour scan speed had a significant impact on the low-cycle fatigue life and surface roughness. Increasing the scanning speed caused the roughness (R_a) to increase, leading to a decrease in the fatigue life. Furthermore, R_v was found to be more noticeable in LCF than R_a [44]. These results are illustrated in Figure 2.21.

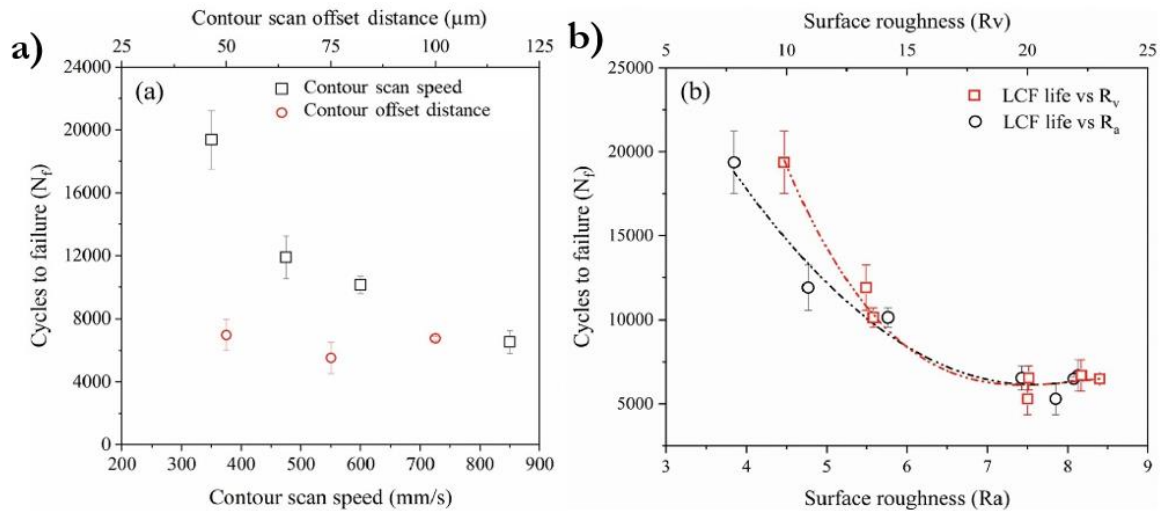


Figure 2.21. Fatigue lifetime as a function of a) contour parameters and b) roughness [44].

A different take on this approach consists in using a contour scan in combination with different surface post-treatments and evaluating their impact on the roughness and subsequent fatigue performance. An investigation using Scalmalloy ® revealed that while the addition of a contour scan leads to a decrease in roughness an increase in fatigue lifetime of 10 – 20 %, the addition of a post-processing step such as shot peening and turning increases the fatigue lifetime by a factor

greater than 2. In the cases without post-processing, the lower roughness level was offset by the emergence of subsurface porosities close to the surface, which have been identified as sites where cracks nucleated. Interestingly, the surface roughness after shot peening was 8 times higher than after turning, yet the comparable fatigue performance was attributed to the emergence of residual compressive stresses at the surface [45].

The impact of surface quality on the fatigue life of $AlSi_{10}Mg$ has been widely investigated. An initial assessment into the fatigue properties of net-shaped and machined samples revealed that net-shaped samples were more sensitive to the surface quality for the same set of process parameters, even when different polishing mechanisms and contouring strategies had been applied. In all cases, machined samples performed better than net-shaped samples, owing to a reduced surface roughness [9]. Further investigations into the bending fatigue of the same alloy for different contour parameters reveal that, by using contour parameters close to the keyhole regime, roughness was diminished, which produced an improvement in fatigue life. However, polished samples still performed better than as-built samples [11].

Subsequent investigations into post-processing and surface treatments of $AlSi_{10}Mg$ have highlighted the impact of surface quality. In general, any surface treatment that reduces roughness and induces compressive residual stresses is able to significantly increase the fatigue life. An investigation into the uniaxial fatigue life under tension-tension conditions of this alloy revealed that at low stress levels, sandblasting produced fatigue performances comparable to those of machining. This surface post-treatment was able to reduce surface roughness by as much as 50 % compared to AB conditions and induced a change in the stress distribution at the material surface, from tensile stresses after L-PBF to compressive stresses [46]. Heat treatments such as T6 in combination with shot peening, laser polishing, machining and grinding have successfully increased the fatigue life of components by 20 – 30 %, as illustrated in Figure 2.22 [47].

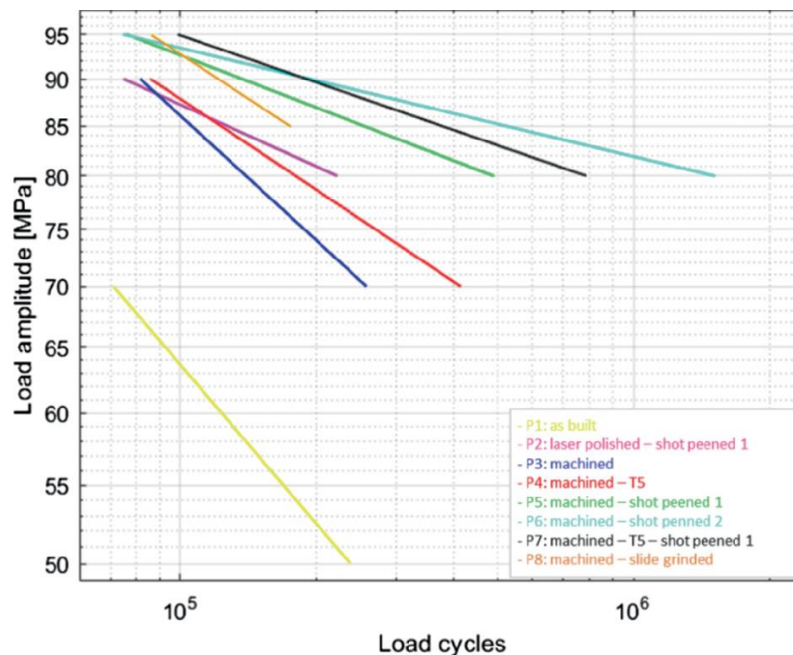


Figure 2.22. Comparison of Wöhler curve for $AlSi_{10}Mg$ samples with different surface treatments [47].

3. Experimental Methodology

3.1. Sample Manufacturing by L-PBF

3.1.1. L-PBF Equipment

The machine used to print all the samples is a ProX ® DMP 200 model manufactured by 3D Systems, property of the IMAP division of the IMMC, Université Catholique de Louvain. Table 3.1 summarizes the most relevant technical information.

Printing Zone (L x W x H) [mm³]	140 x 140 x 100
Laser Beam Wavelength [nm]	1070
Maximum Laser Power [W]	273.6
Shielding Gas	Argon
Oxygen Limit [ppm]	500

Table 3.1. Technical information about the ProX ® DMP 200 L-PBF machine.

3.1.2. Powder Characteristics and Preparation

The manufacturing of samples was carried out using AlSi₁₀Mg recycled metallic powder produced by TLS Technik. According to the manufacturer, the particle size distribution ranges from 15 to 53 µm, and the particle morphology is spherical. The chemical composition was analyzed using inductively coupled plasma (ICP) at UCLouvain and is presented in Table 3.2.

Element	Al	Si	Mg	Fe	Ga	Ti	Ni
Mass Fraction (%)	89.12	10.02	0.32	0.11	0.01	0.01	<0.01

Table 3.2. Chemical composition of AlSi10Mg powder.

In order to recycle the powder and remove large particles and conglomerates of particles, manual sieving with a screen of 63 µm was carried out at frequencies ranging from 30-50 Hz before each printing. The sieving time varied from 12 to 15 minutes, with average yields of approximately 70 grams of powder per sift. Sieving of recycled powder was carried out after each printing.

3.1.3. Specimen Geometry

Cylindrical tensile and fatigue specimens were printed, and their geometry defined according to the ASTM E8/E8M-13a [48] and ASTM E466-15 [49] respectively. Two different fatigue specimen geometries, illustrated in Figure 3.1, labeled Type A and Type B respectively, were used. Half-samples, illustrated presented in Figure 3.3, were used in the optimization and selection process of geometry and process parameters in order to ensure the presence of keyholes and an improved surface roughness, as discussed in Section 4.1. The tensile specimen geometry is illustrated in Figure 3.2. All dimensions are in mm.

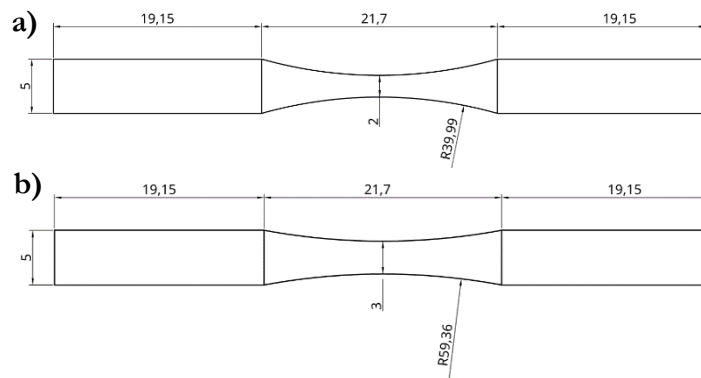


Figure 3.1. Cylindrical a) Type A and b) Type B fatigue specimen geometry.

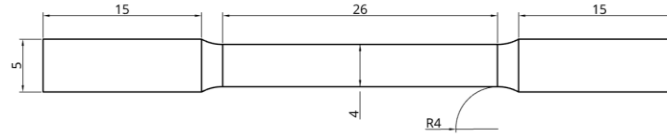


Figure 3.2. Cylindrical tensile specimen geometry.

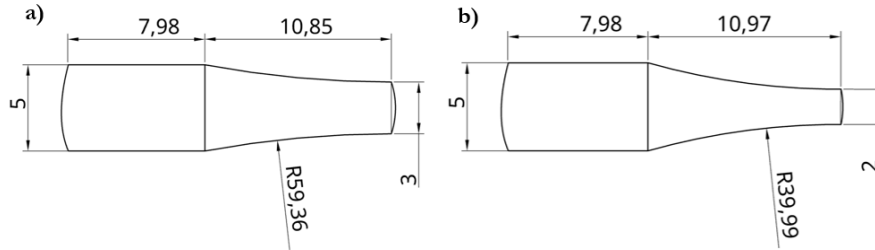


Figure 3.3. Geometry of a) Type B and b) Type A half-samples used in the optimization and selection process.

3.1.4. Printing Parameters and Scanning Strategies

- **Scanning Strategies.** Two different scanning strategies were used in the course of this investigation: no contour and contour first. Even though none of the samples printed without contours were tested, an illustration of these two strategies is presented in Figure 3.4 for completeness.

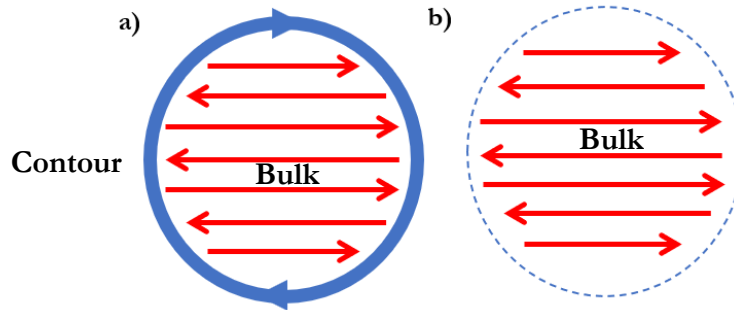


Figure 3.4. a) Contour first and b) No contour scanning strategies.

- **Bulk Parameters.** The bulk parameters were fixed throughout the course of this investigation for all samples in all batches. These are optimized to produce the maximum density in the core of the sample and were used by Mache in a previous investigation [11]. The layer thickness and hatching space were selected based on the results of Poncelet et al. [10], as they produced the best roughness. These are summarized in Table 3.3.

Laser Power [% of Maximum]	85
Scanning Speed [mm/s]	1200
Jump Speed [mm/s]	5000
Defocus [mm/10]	0
Hatching Space [μm]	100
Offset (from STL) [μm]	-90
Layer Thickness [μm]	30

Table 3.3. Fixed bulk parameters.

- **Contour Parameters.** All contour parameters, except scanning speed and laser power, were fixed throughout the course of this investigation for all samples in all batches. These are summarized in Table 3.4. The scanning speeds used in the contour range from 200 to 700 mm/s, depending on the batch, which will be detailed in Annex 10.1.

Laser Power [% of Maximum] (Bulk/Contour)	85/100
Jump Speed [mm/s]	5000
Defocus [mm/10]	0
Offset (from STL) [μm]	-30

Table 3.4. Fixed contour parameters.

A full description of each batch can be found in Annex 10.2.

3.2. Defect Characterization

3.2.1. Keyhole Porosities

Preliminary observations of keyhole porosities were carried out in order to validate their presence along the contours of the printed samples during the optimization and selection stage, as discussed in Section 4.1. The sample preparation procedure for microscopic observation was the following:

- ✓ Samples were cut across the cross section using a hand saw at two locations: the head and the core sections, as illustrated in Figure 3.5, in order to extract two subsections of material.

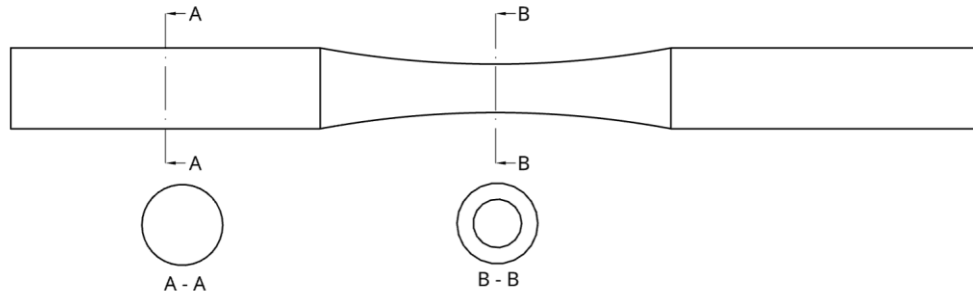


Figure 3.5. Head and core cross-sectional view locations for preliminary keyhole observation.

- ✓ The extracted subsections of material were mounted in non-conductive resin at 10 kN, with a heating time of 6 minutes and a cooling time of 3 minutes.
- ✓ Mounted samples were polished to a mirror-like finish in order to carry out microscopic observations of the head and core sections. Table 3.5 summarizes the information of the different polishing steps. Grinding and polishing were carried out using a Forcipol 102 Metkon and Struers DP metallographic grinding and polishing machines, respectively.

Step	Operation	Abrasive Material	Grade / Grit	Time [min]	Speed [RPM]
1	Grinding	Silicon Carbide Sandpaper	320	5	190
2	Grinding	Silicon Carbide Sandpaper	1200	5	190
3	Polishing	Polycrystalline Diamond Paste	3 μm	15	150
4	Polishing	Polycrystalline Diamond Paste	1 μm	15	150

Table 3.5. Grinding and polishing steps for mounted samples.

- ✓ Chemical etching of the polished surface was carried out by immersing the mounted samples in a Keller reagent, containing 150 ml of water, 5 ml of nitric acid, 7 ml of chlorohydric acid, and 7 ml of hydrofluoric acid for 30 seconds, followed by neutralization using distilled water by immersion for 2 minutes. The purpose of this etching was to enhance the contrast of the microstructural features.

After the samples were properly prepared, a Keyence VHS 7000 series 3D digital microscope was used to observe and validate the keyhole porosities in the contour regions. Coaxial lighting coupled with a 40x magnification were used to observe and capture the images of the cross sections. An in-depth characterization on the distribution and volume of porosities was carried out using X-Ray tomography. To carry it out, an EasyTom XL CT scanner was used, with a voxel size of 6 μm . Processing of the tomography data was done using the Avizo 2023.1 software and included the conversion from 16 to 8-bit as well as the use of a bilateral and unsharp masking filters.

3.2.2. Surface Roughness

Roughness was characterized using the 3D serial image recording and composition module coupled with line roughness measurements of the Keyence VHS 7000 series 3D digital microscope. The procedure was based on the results by Poncelet et al. [10] and the assessments on the impact of the different profile measurement parameters by Nagalingam et al. [50]. These measurements were carried out in the core and head of the samples, These parameters are the following:

- **λ_s** : Low-pass filter that eliminates the smallest scale elements from the surface.
- **λ_c** : High-pass filter that eliminates waviness from the surface, allowing for the extraction of only the roughness components. It is also known as the sampling length.
- **L_n** : Assessment length over which the roughness is averaged.
$$L_n = 5\lambda_c$$
- **L_t** : Sampling length that considers a distance before and after the assessment length.

$$L_t = 5\lambda_c + \left(\frac{\lambda_c}{2}\right)_{pre} + \left(\frac{\lambda_c}{2}\right)_{post}$$

Following the guidelines of the EN ISO 4288 concerning laboratory setups for surface roughness measurements [51], the parameters selected are shown in Table 3.6

λ_s [μm]	8
λ_c [mm]	2.5
L_n [mm]	12.5
L_t [mm]	15

Table 3.6. Profile roughness measurement parameters

To perform the roughness measurements, a 3D horizontal traveling serial recording was performed over the section of interest with a magnification of 150x. The highest and lowest points for the lens focusing were defined at the head and core sections, respectively. Three traveling lengths were defined for the head and the core sections as illustrated in Figure 3.6, and the average of the three roughness measurements was taken as the actual roughness measurement.

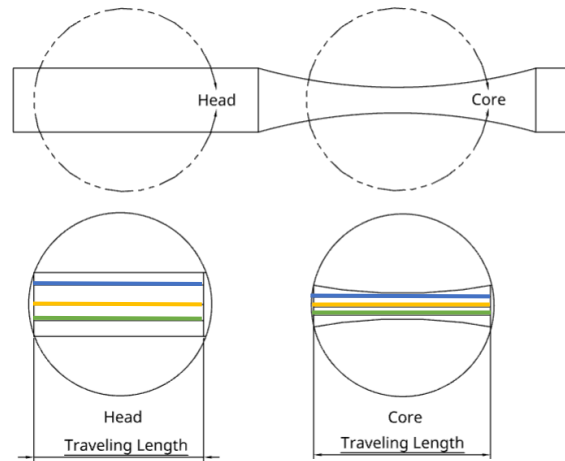


Figure 3.6. Description of profile roughness measurement procedure.

3.3. Stress Relief Heat Treatment

A stress relief heat treatment (SRHT) was carried out in all the fatigue samples and half of the tensile samples to achieve residual stress relaxation without significantly altering the microstructure and mechanical properties. To achieve this, the conditions employed by [52] were replicated. Thus, in order to remain as close to the as-built (AB) conditions, the decision was made to perform the heat treatment at 250 °C for 2 hours. The heat treatment was carried out using a Heraeus VT 5050 EKP oven, with a maximum working temperature of 400 °C. The procedure was the following:

- ✓ Samples were inserted into the oven at room temperature (20°C).
- ✓ In order to reduce the oxygen concentration in the oven, the air was pumped out until a pressure of 200 mbar was reached, after which the oven was pressurized with Argon until a pressure of 600 mbar was reached. The procedure was repeated three times.
- ✓ After the three Argon flushes, the oven was pressurized to 400 mbar.
- ✓ The target temperature was set at 250 °C. Heating until the setpoint took 25 minutes. When the target temperature was reached, samples were treated for two hours.
- ✓ After two hours, the oven was turned off, and samples were left to cool overnight.

3.4. Mechanical Testing

3.4.1. Static Tensile Testing

Tensile tests consist of the application of a tension load on a material sample by prescribing a load or displacement. These tests are performed using universal testing machines, which include a double-column frame, a load cell, testing software, application-specific grips, and a strain-measuring device (strain gauge, extensometer, etc.). In order to properly characterize the mechanical behavior, the specimen geometry needs to ensure a uniform stress field away from the grips and avoid geometrical effects such as stress concentration. The tensile test was carried out using a Zwick/Roell Instron® testing machine with a pre-charge of 5 N and an extension rate of 1 mm/min, in combination with an extensometer with 20 mm of gauge length. The setup of the sample is shown in Figure 3.7a.

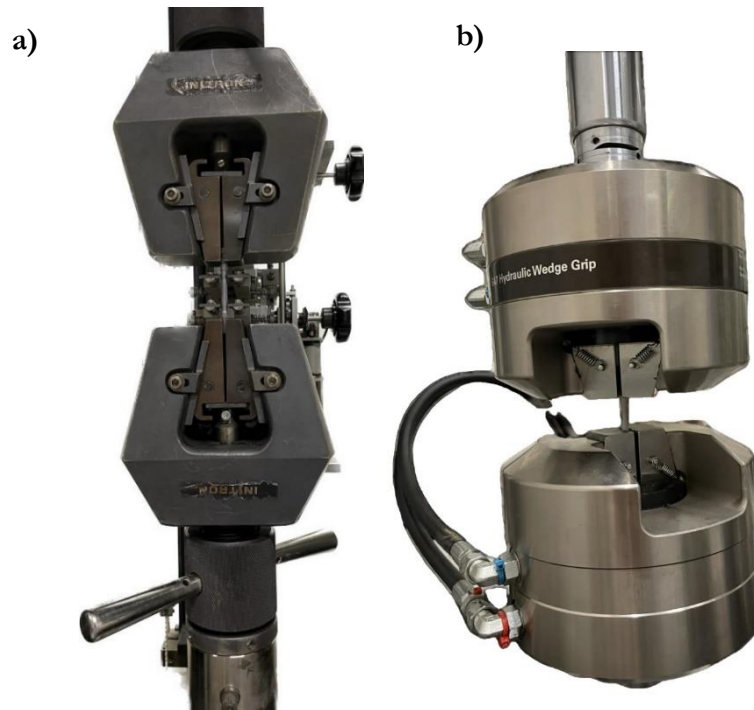


Figure 3.7. Experimental setup of the a) tensile and b) fatigue tests.

The procedure to perform the tensile tests was the following:

- ✓ The samples' cross section were measured using calipers in three locations, and the results averaged.
- ✓ The grips were unfastened, and the force was set to zero. The sample to be tested was placed in the grips, and these were slightly tightened.
- ✓ The extensometer's position was tested. If adequate, the grips were securely tightened.
- ✓ The sample designation and diameter were input into the software, and the test launched.

The results were exported into Microsoft Excel sheets and processed using a custom Python script. The mechanical properties and parameters calculated and extracted from this data were:

- **Nominal Stress.** The nominal stress refers to the force applied over the cross-sectional area of the undeformed sample. The equation used to calculate it is:

$$\sigma_{\text{nom}} = \frac{F}{S_0}$$

- **Nominal Strain.** The nominal strain refers to the change of length divided by the length of the undeformed sample. The equation used to calculate it is:

$$\epsilon_{\text{nom}} = \frac{\Delta l}{l_0}$$

- **True Stress.** The true stress refers to the force applied over the actual or deformed cross-sectional area of the sample. The equation used to calculate it is:

$$\sigma_{\text{true}} = \sigma_{\text{nom}}(1 + \epsilon_{\text{nom}})$$

- **True Strain.** The true strain refers to the deformation of the sample. The equation used to calculate it is:

$$\epsilon_{\text{true}} = \ln(1 + \epsilon_{\text{nom}})$$

- **Young's Modulus.** The Young's or elastic modulus is a mechanical property of isotropic materials that measures the material's stiffness when subjected to uniaxial tension or compression. To calculate it, a calculation range was defined with lower and upper stress limits of 0 and 150 MPa respectively. These limits were based on previous stress-strain curves for AlSi₁₀Mg. The data within this range was subjected to a linear regression, and the resulting slope corresponded to the Young's modulus.
- **Yield Stress.** The yield stress refers to the minimum stress at which a material will experience plastic deformations without an increase in the applied load and marks the end of the linear elastic regime. To calculate it, the 0.002 offset method was used.
- **Fracture Strain.** The fracture strain refers to the ultimate strain the material reached before failure of the specimen. It was designated as the last value of fracture strain before the load level dropped abruptly.

3.4.2. Fatigue Testing

Fatigue tests consist of the application of a cyclical force or displacement within predetermined levels at a specified testing frequency until a specific number of cycles has been reached or the material fails by fracture. This test is usually performed such that the stress never exceeds the yield stress. However, in low-cycle fatigue tests, the yield stress may be exceeded, opening the possibility for alternating plasticity to occur. Unlike tensile tests, there are many possible equipment alternatives which can be used to conduct a fatigue test depending on the type of stress and loading condition which is under investigation. Some of the most common fatigue tests include alternating rotatory bending fatigue and uniaxial tension fatigue. Factors such as surface condition and geometry can have a significant impact on the results.

The fatigue tests were carried out using an MTS Bionix Tabletop System, with a maximum rated force of 25 kN. Samples were tested under tension-tension conditions (setup illustrated in Figure 3.7b) since it has been reported that AlSi₁₀Mg is better under uniaxial tension-compression [26]. The test was set up with a stress ratio of $R=0.1$ and a frequency of 30 Hz. The procedure to perform the fatigue tests was the following:

- ✓ Fatigue samples were tested in three conditions: heat treated (SRHT), sand blasted (SHRT + SB), and polished (SHRT + P). Sandblasting was carried out for three minutes using a Laborex Type 1250 industrial cleaning and finishing machine. Table 3.7 summarizes the information of the different polishing steps and equipment used. For circumferential grinding, a Myford Lathe with a set rotation speed of 150 rpm was used. Longitudinal grinding to eliminate detrimental circumferential scratches was done by hand using cotton.

Step	Operation	Abrasive Material	Grit	Time [min]	Method	Direction
1	Grinding	Silicon Carbide Sandpaper	1200	5	Lathe	Circumferential
2	Grinding	Silicon Carbide Sandpaper	2400	5	Lathe	Circumferential
3	Grinding	Silicon Carbide Sandpaper	4000	5	Lathe	Circumferential
4	Polishing	Polycrystalline Diamond Paste	6 μm	5	Manual	Longitudinal
5	Polishing	Polycrystalline Diamond Paste	3 μm	5	Manual	Longitudinal
6	Polishing	Polycrystalline Diamond Paste	1 μm	5	Manual	Longitudinal
7	Polishing	Polycrystalline Diamond Paste	1/4 μm	5	Manual	Longitudinal

Table 3.7. Grinding and polishing steps for fatigue specimens.

- ✓ The diameter of the samples was measured using an optical comparator. Two measurements at the narrowest cross-section were performed, and the average was taken as the true diameter of the specimen.
- ✓ Based on the stress ratio, the diameter of the specimen, and the target stress level, the maximum and minimum force were determined. A summary of the fatigue campaign information is presented in Table 3.8 , and a full description can be found in Annex 10.3.

Contour Parameters		Stress Range [MPa]	Surface Condition
Laser Power [%]	Scanning Speed [mm/s]		
100	300	220-140	SRHT
			SRHT+P
			SRHT + SB
	600	220-140	SRHT
			ST+P
			SHRT + SB
85	1200	220-140	SHRT

Table 3.8. Summary of fatigue testing campaign.

Processing of the fatigue test information was done by the creation of a classical and a yield stress-normalized Wohler's curve.

3.5. Fracture Surface Analysis (FSA)

A fractography of the tensile and fatigue specimens was carried out in order to determine the failure mechanisms and the most important characteristics of the fracture surface. To do this, the preliminary approach used the 3D fracture surface composition using the Keyence VHS 700 Series digital microscope in order to identify the overall characteristics of the samples. Afterwards, selected samples exhibiting special features were observed more in depth using the FEGSEM ULTRA 55 scanning electron microscopes under progressive magnifications of 100x, 500x, and 2000x.

4. Results

This section presents the results of the optimization and selection of the process parameters and the mechanical (static tensile and uniaxial fatigue) tests. The fracture surface and cross-sectional views are used to show microstructural differences, the presence of porosities, and sites of crack nucleation and propagation. For the sake of clarity, the diagram in Figure 4.1 illustrates the core and head sections of the specimens as well as the contour and bulk regions in each sample.

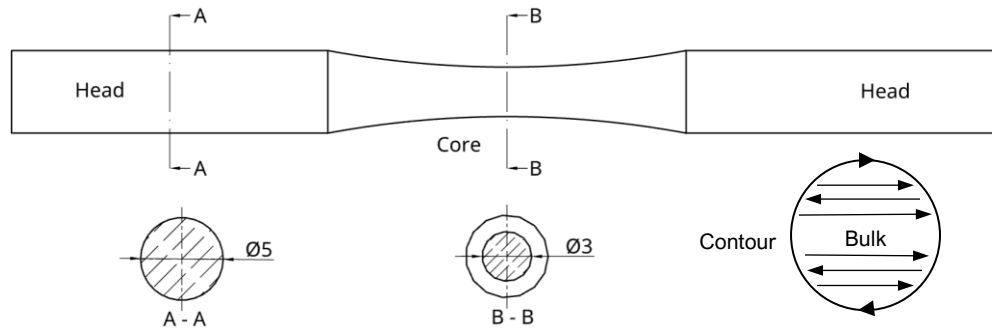


Figure 4.1. Illustration of the core/head sections and contour/bulk regions in each fatigue specimen.

4.1. Optimization and Selection of Geometry and Process Parameters

Based on the lack of keyhole porosities and high roughness observed in the Test Batch (TB) (see details in Annex 10.1), an Optimization Batch (OB) consisted of a systematic study varying the core diameter and the scanning speed of contours on fatigue half-samples. The purpose of this process was to determine the scanning speed and core diameter that produce low roughness (comparable to those in the literature) and keyhole porosities in the contour region of the core. As evidence in Figure 4.2, when a core diameter of 2 mm is used, keyhole porosities are not observed in speeds above 200 mm/s. It is possible to see that even at 200 mm/s, the keyhole porosities in the contour of the core region are small in size compared to the melt pools and are not evenly distributed along the contour. In contrast, Figure 4.3 shows that for a core diameter of 3 mm, keyhole porosities appear along the entire length of the contour in the range of 200 – 400 mm/s. In both cases, the gradual disappearance of keyhole porosities in the contour as the scanning speed is expected, as the TED is decreasing from 1368 J/m (at 200 mm/s) to 390 J/m (at 700 mm/s). However, despite the fact that for both diameters the TED is the same, the lack of keyholes for the smaller diameter suggests that the heat transfer mechanism and thermal profile along the contour is not the same. This is discussed further in Section 5.2.

The effect of the geometrical change for varying scanning speeds is also evident in the case of the surface roughness (R_a) exhibited by the samples, as evidenced by Figure 4.4. Compared to the reference (contour parameters set to be the same as those in the bulk 232.5 W, 1200 mm/s, on the extreme left of the graph), the trend indicates that for an increasing TED, the roughness diminishes for both diameters. However, the improvement in surface quality is higher for the samples with a 3 mm diameter. The reference samples with 2 and 3 mm diameters have a comparable surface roughness of 8.69 and 8.48 μm respectively. However, as the TED increases, the improvements in surface quality range from 27.5 to 47.5 % for the speed range 200 – 700 mm/s. Furthermore, whereas the best roughness obtained in the 2 mm samples was 5.77 μm at 400 mm/s, all the roughness measurements in the 3 mm samples in the range 200 – 700 mm/s are below 5 μm , with the optimum roughness being 3.81 μm at 400 mm/s. The error in the roughness data decreases significantly for the 3 mm samples compared to the 2 mm samples.

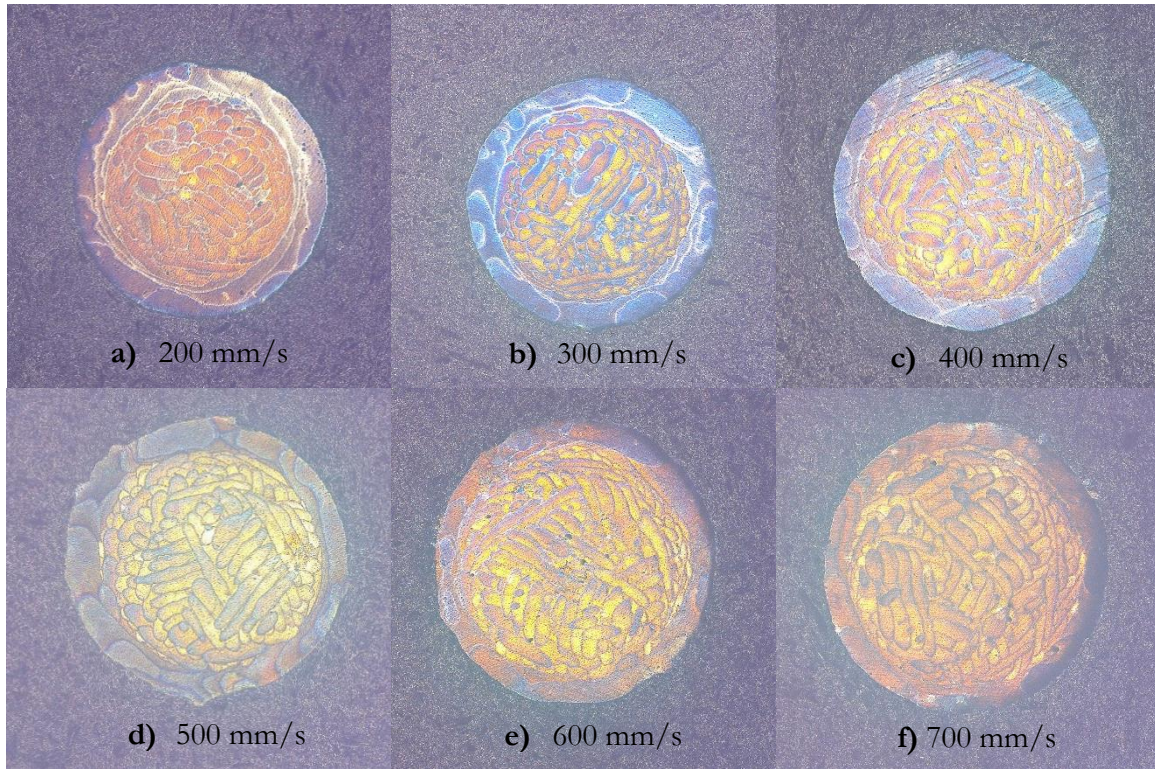


Figure 4.2. Effect of scanning speed on the contour keyhole porosities of 2 mm diameter half samples.

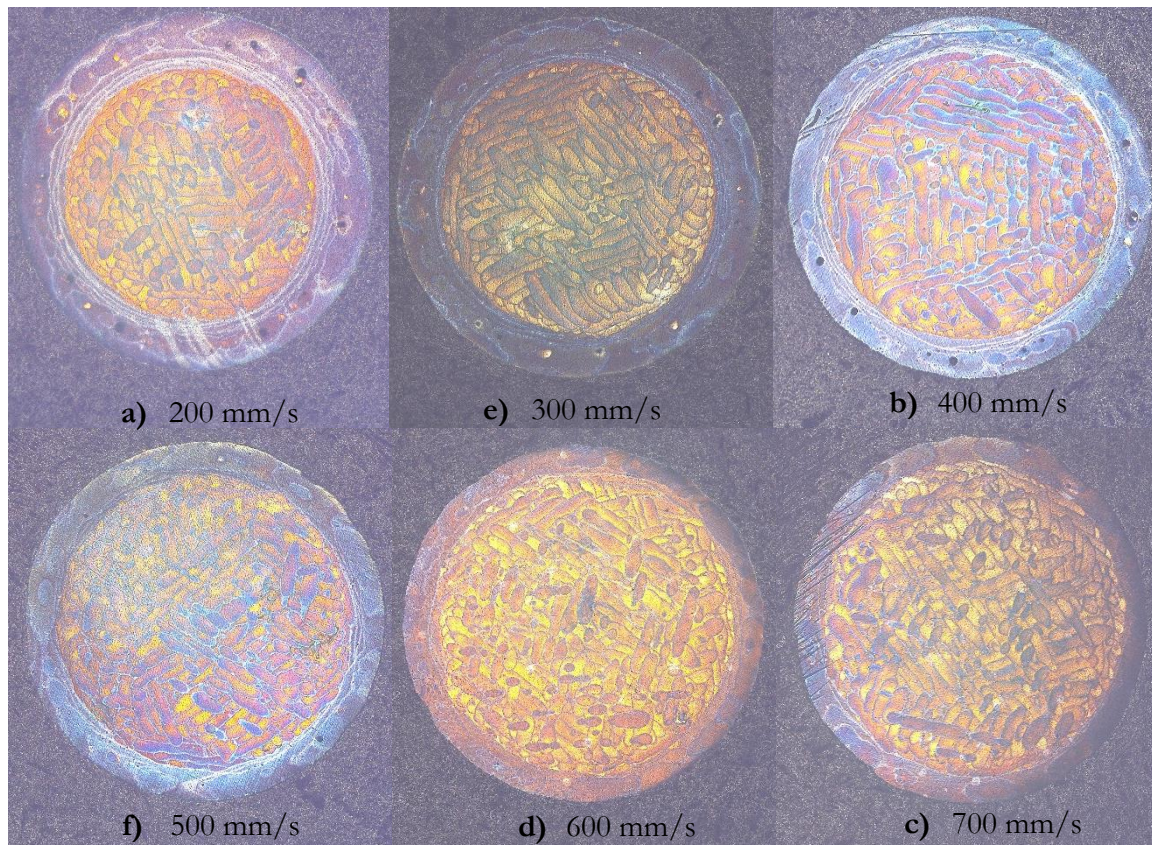


Figure 4.3. Effect of scanning speed on the contour keyhole porosities of 3 mm diameter half samples.

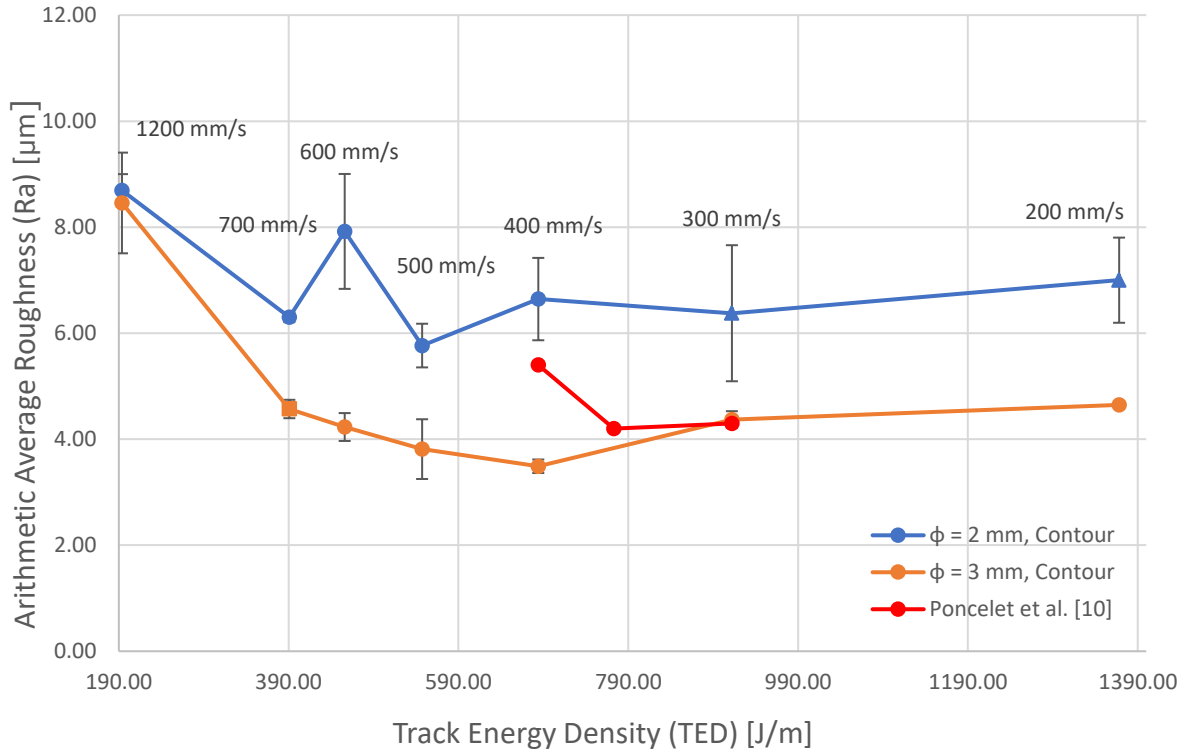


Figure 4.4. Effect of scanning speed on the surface roughness of 2 and 3 mm diameter half samples.

4.2. Overview of Surface Roughness and Porosities

From the results of the OB, the 3 mm core diameter was selected. Scanning speeds of 300, 600, and 1200 mm/s were chosen for the “hot” (keyholes and low roughness), “cold” (no keyholes and low roughness), and reference contour cases, respectively. The process parameters used for the Specimen Batches 3 and 4 (SB3 and SB4) are detailed in Table 4.1.

Sample Designation	Bulk			Contour		
	Laser Power [W]	Scanning Speed [mm/s]	TED [J/m]	Laser Power [W]	Scanning Speed [mm/s]	TED [J/m]
100-300 (Hot Contour)	232.5	1200	193.8	273.6	300	912
100-600 (Cold Contour)					600	456
85-1200 (Reference)				232.5	1200	193.8

Table 4.1. TED of the “hot” contour, “cold” contour, and reference cases for SB3 and SB4.

An in-depth porosity characterization revealed that keyhole porosities in the contour of the 100-300 samples vary in size. As evidenced in Figure 4.5a, the largest keyholes (white arrows) with a diameter range of approximately 50 – 100 μm are located in the central region of the contour melt pool. However, a secondary set of smaller porosities (green arrows) was observed near the outer and inner edges of the contour melt pool, which are under 10 μm in size. Some irregular porosities (red arrows), whose shape suggests lack of fusion, were present in the bulk region of the sample, despite the fact that density-optimized parameters were selected for the bulk. A high-resolution CT scan was carried out in order to highlight the presence of these microporosities in the contour, as illustrated in Figure 4.5b, which consists of a two-dimensional projection of 450 slices (900 μm thickness) at the core of the sample.

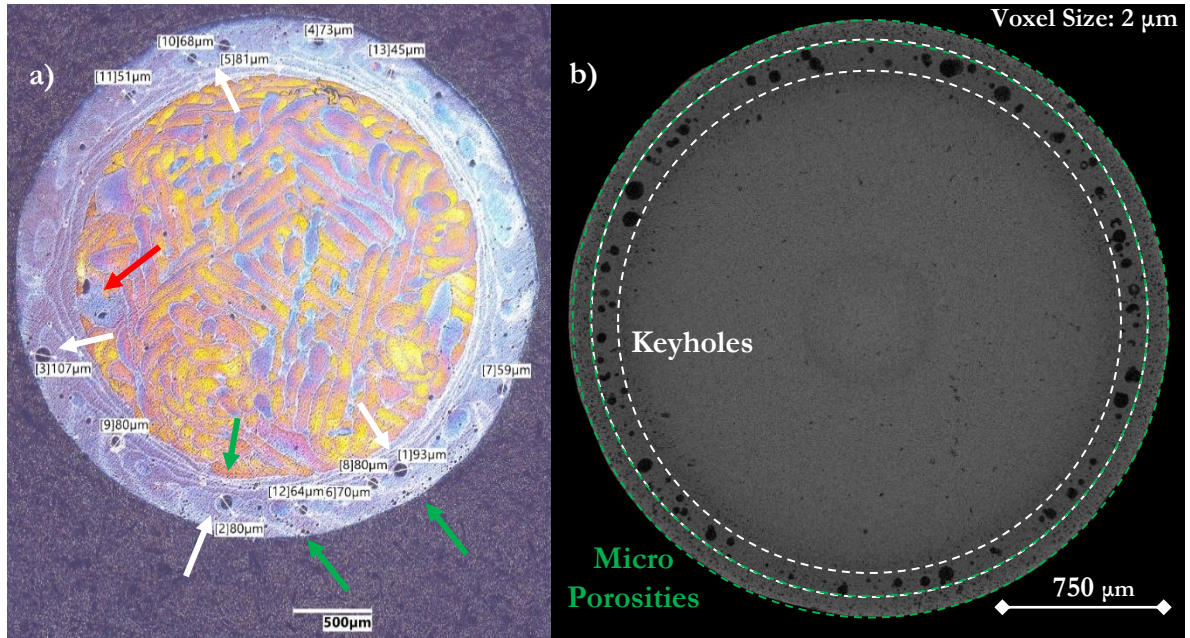


Figure 4.5. a) Microscopy and b) CT scan characterization of contour and bulk porosities in 100-300 sample.

Additional CT scans of the whole samples provided more insights into the nature and distribution of the defects in the 100-300 and 100-600 specimens. As evidenced in Figure 4.6a and b, keyholes are evenly distributed along most of the contour of the core region of the 100-300 samples, and few bulk porosities are visible. The head section of the sample presents more porosities in the contour than the core (the reason for which will be addressed in Section 5.2). Furthermore, the head of the sample exhibits more porosities in the bulk section which are small in size compared to those in the contour. The microporosities in the inner and outer edges of the contour are not visible in these scans due to the voxel size and corresponding resolution used in the scanning process. However, these images capture the larger keyholes which account for the majority of the contour porosity.

Unlike the 100-300 samples, the 100-600 samples do not exhibit significant keyholes in the contour and little porosities in the bulk at the core region, as shown in Figure 4.6c and d. The absence of contour keyholes is due to the TED, approximately 1.75 times lower than in the 100-300 samples. In the head of the samples, some contour porosities are present, which could be attributed to the geometry effect observed during the optimization process. However, the head also exhibits an inordinately large number of small, apparently round porosities in the bulk. These porosities appear to be of the same order of magnitude as those observed in the contour. Since both the 100-300 and 100-600 samples were manufactured with the same bulk parameters, the TED alone cannot account for this behavior. Though the 100-300 samples do exhibit more bulk porosities compared to the core, they do not exhibit as much porosities as the 100-600 samples.

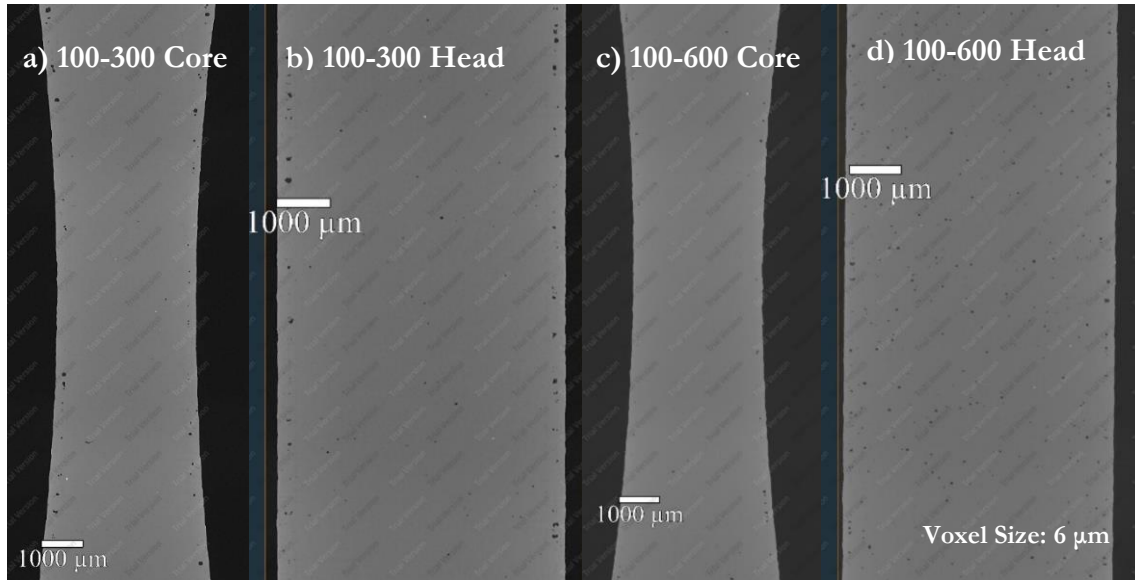


Figure 4.6. Longitudinal section of a 100-300 at the core (a) and head (b) and of a 100-600 sample at the core (c) and head (d).

In order to ensure that the porosity results were not an isolated occurrence, an observation into the cross-sections of the heads of tested 100-300 and 100-600 fatigue specimens was carried out. In all cases, the higher frequency of bulk porosities in the heads of the 100-600 samples compared to the 100-300 samples was confirmed, as illustrated in Figure 4.7. The porosities appear to have a round morphology, which indicates they may be keyholes. The dents on the contour are due to the grips of the fatigue machine.

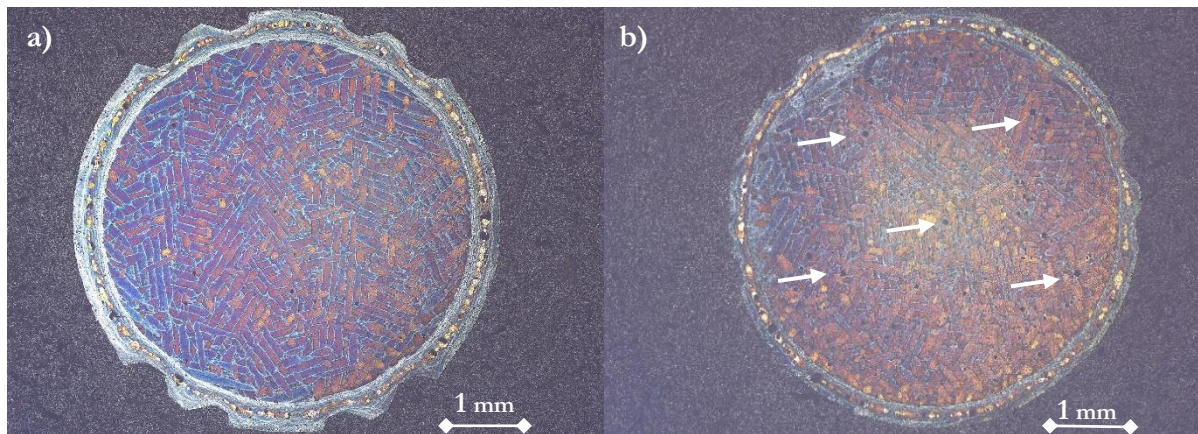


Figure 4.7. Cross-sectional view of the contour and bulk at the head of a) 100-300 and b) 100-600 samples.

The increase in TED in the contour of the samples causes a marked decrease in the lateral surface roughness (R_a) compared to the reference contour in 85-1200 samples, as evidenced in Figure 4.8. Compared to a reference value of $8.82 \mu\text{m}$, reductions of 52 % and 41 % for the 100-300 and 100-600 samples, respectively, were observed. Compared to the results of the OB, the surface roughness (R_a) of the full samples contrasted to that of the half samples is comparable despite the differences in thermal history. In the half samples, roughness values of $4.37 \mu\text{m}$ and $4.23 \mu\text{m}$ were obtained for the 100-300 and 100-600 samples, respectively. In the full samples, roughness values across both batches averaged $4.24 \mu\text{m}$ and $5.23 \mu\text{m}$ for the 100-300 and 100-600 samples, respectively. Therefore, no significant differences among batches were observed.

The maximum profile roughness (R_z) showed a trend similar to that of the average surface roughness (R_a), as shown in Figure 4.9. In this case, increasing the TED caused R_z to drop from an average of $55.67 \mu\text{m}$ to $22.99 \mu\text{m}$ and $29.05 \mu\text{m}$ for the 100-300 and 100-600 samples, which represents an improvement of 58 % and 47 % respectively. This indicates that increasing the contour TED has a larger impact on the maximum profile roughness than the average arithmetic roughness.

Figure 4.10 illustrates how the contributions of the peak (R_p) and valley (R_v) roughness to the maximum profile roughness evolve with TED. In the 85-1200 samples, R_p constitutes 65% of the total R_z , while R_v accounts for 45% of R_z . Increasing the TED reduces both R_p and R_v and causes the overall reduction in R_z described above, but this reduction is not proportional for R_v and R_p . While R_p is reduced from $32.9 \mu\text{m}$ to $11.7 \mu\text{m}$ and $7.9 \mu\text{m}$, R_v is reduced from $26.9 \mu\text{m}$ to $18.2 \mu\text{m}$ and $16.5 \mu\text{m}$ in the 100-600 and 100-300 samples respectively. As a result, R_v displaces R_p as the main contributor to R_z , and accounts for 61% and 68% of the total R_z in the 100-300 and 100-600 samples. From a physical point of view, increasing the TED has a greater impact on the attached un-melted particles than on the grooves that exist on the surface of the samples. This will be discussed further in Section 5.1.

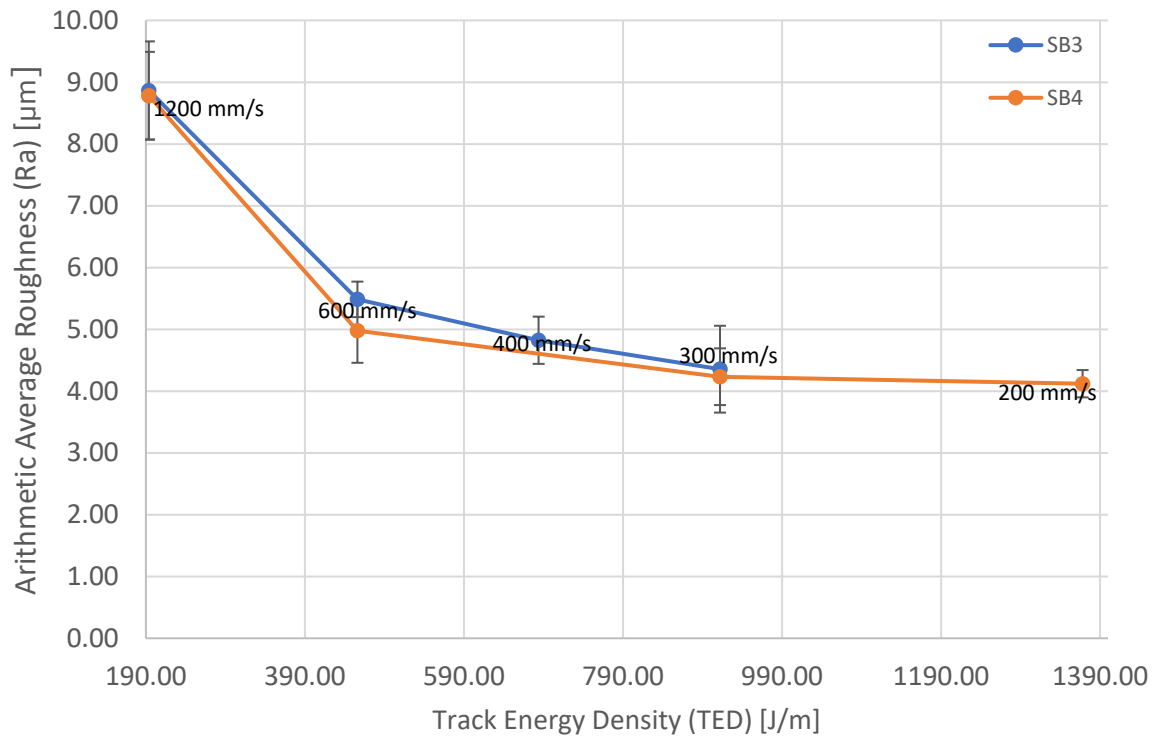


Figure 4.8. Effect of TED on the average arithmetic roughness (R_a) of SB3 and SB4.

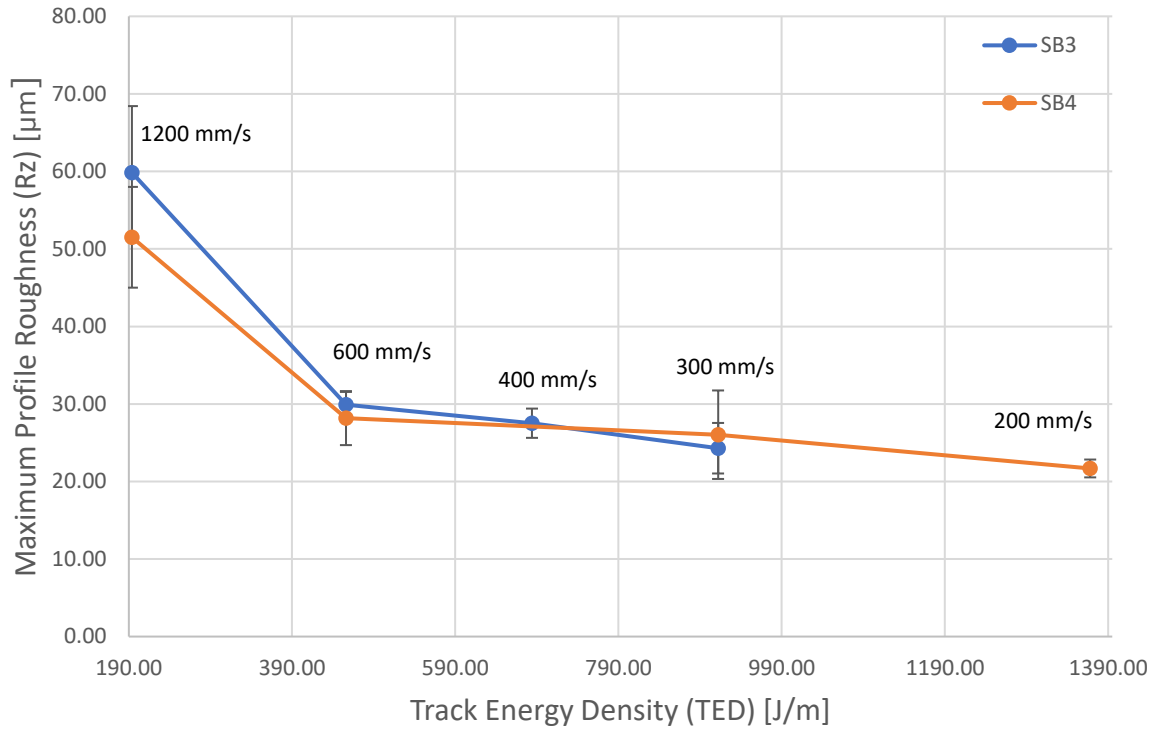


Figure 4.9. Effect of TED on the maximum profile roughness (R_z) of SB3 and SB4.

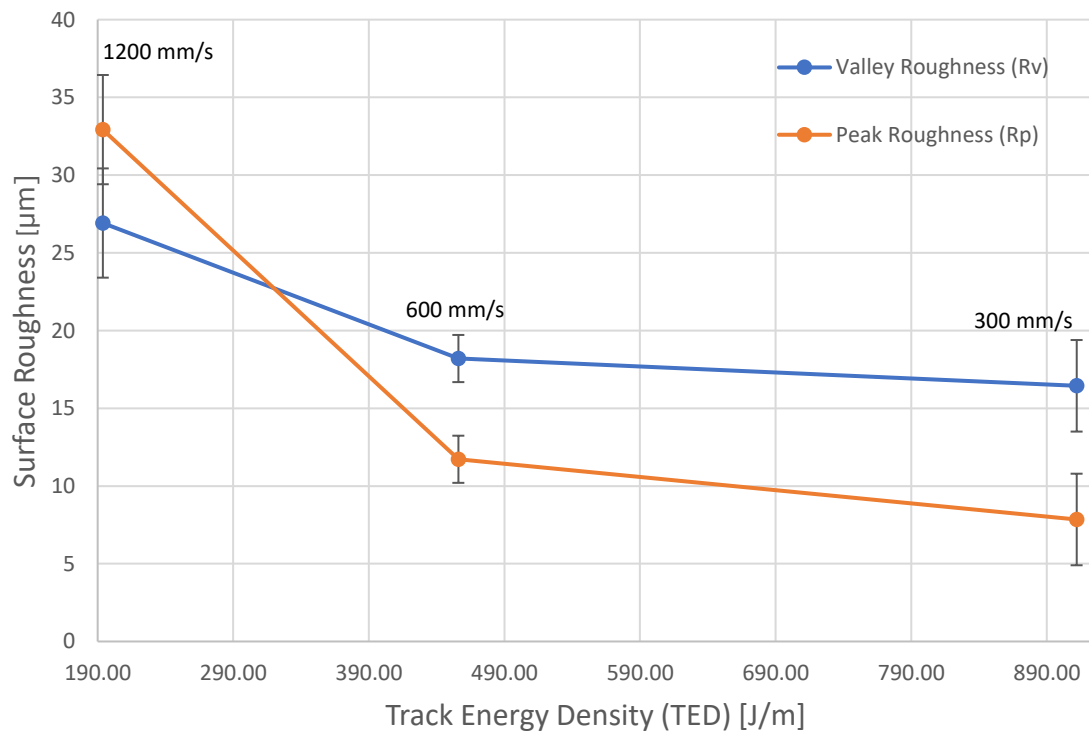


Figure 4.10. Effect of TED on the valley (R_v) and peak (R_p) roughness in SB3 and SB4.

4.3. Microstructure

As-built (AB) samples from the 100-300, 100-600, and 85-1200 groups from SB3 and SB4 were subjected to a heat treatment (SRHT) at 250°C for 2 hours in order to relax the residual stresses product of the high cooling rates associated with L-PBF. Figure 4.11 presents a comparison between the microstructure of the AB and SRHT samples at the bulk and at the contour regions in the core of a 100-300 sample.

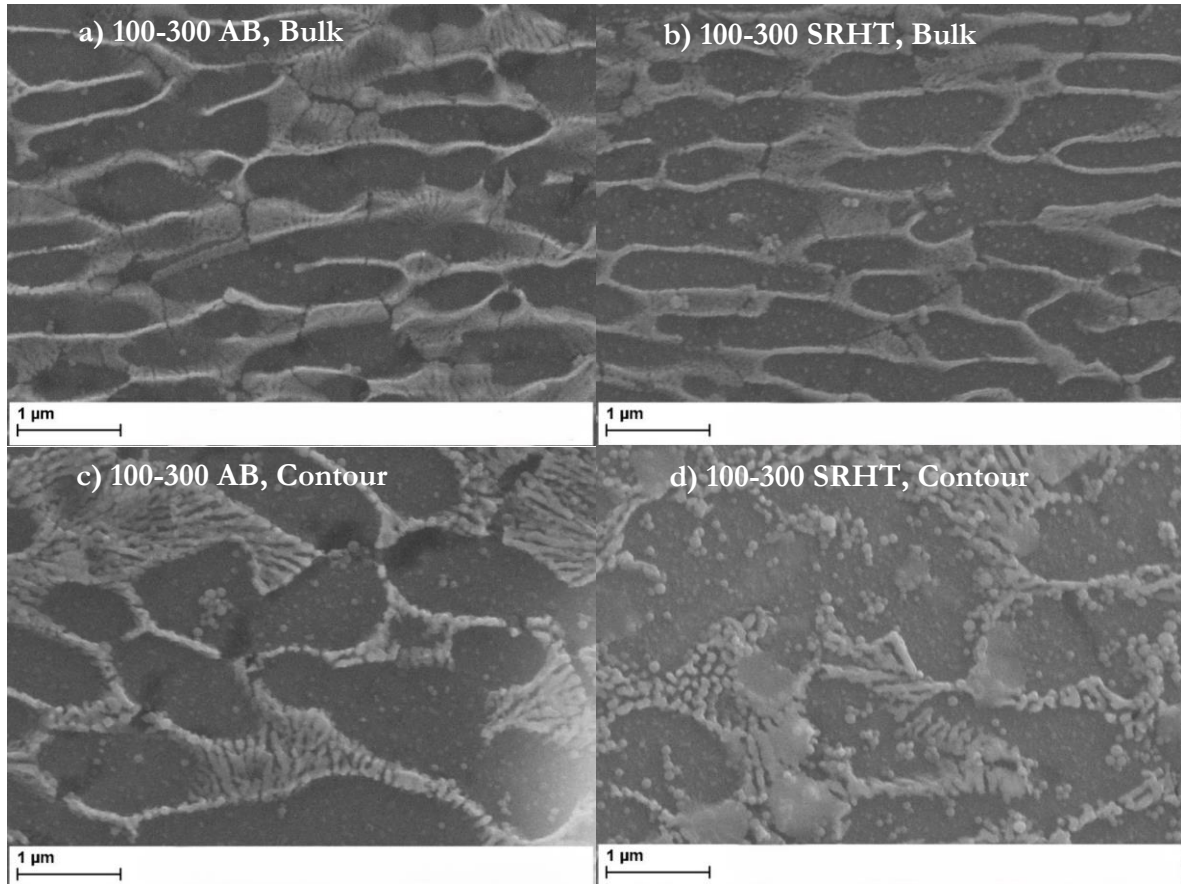


Figure 4.11. Microstructural comparison between an AB (in the bulk (a) and contour (c)) and a SRHT (in the bulk (b) and contour (d)) 100-300 sample.

The picture showcases the effect the heat treatment and contour parameters have on the microstructure. Prior to the heat treatment, the AB sample exhibits a fine Si network in the bulk, while the contour presents a coarser Si network. This suggests that the higher TED used in the contour (912 J/m) induced coarsening of the Si network compared to the lower TED (193 J/m) used in the bulk. The nature of the microstructure in the bulk of the SRHT is similar to that observed in the AB sample, as the network has not been visibly globularized nor degraded. It is interesting to note that more Si precipitates are visible in the Al matrix in the SRHT sample compared to the AB sample. Unlike the bulk, the heat treatment seems to have affected the contour region more significantly. The Si network seems to have undergone more coarsening and globularization in the SRHT sample compared to the AB sample. This indicates that the heat treatment affected the already coarse Si network in the contour more than it affected the fine Si network in the bulk. Overall, no large-scale microstructural changes were observed.

4.4. Static Properties

Sample groups 100-300, 100-600, and 85-1200 in as-built (AB) and heat treated (SRHT) conditions from SB3 and SB4 were subjected to tensile tests, and the resulting stress-strain curves are shown in Figure 4.12. The blue and green curves showcase the differences between SB3 and SB4 in the AB state, while the orange and red curves showcase the same differences in the SRHT state. The average values for the elastic modulus in the build direction, yield stress, ultimate tensile stress (UTS) and fracture strain from SB3 and SB4 are summarized in Table 4.2.

Sample Group	Young's Modulus [GPa]		Yield Stress [MPa]		Ultimate Tensile Stress (UTS) [MPa]		Fracture Strain [-]	
	AB	SRHT	AB	SRHT	AB	SRHT	AB	SRHT
100-300	58.67	70.09	239.55	238.95	431.03	391.95	0.05	0.03
100-600	66.66	65.44	250.87	244.38	426.09*	413.01	0.04*	0.04
85-1200	62.76	66.37	252.79	238.88	456.75	403.78	0.05	0.04

Table 4.2. Summary of average mechanical properties for SB3 and SB4.

The values of the Young's Modulus show no major differences and range from approximately 60 GPa to 70 GPa. Likewise, the values for the yield stress in the AB and SRHT conditions do not show any significant differences and range from approximately 240 MPa to 250 MPa. Unlike the Young's Modulus and yield stress, it is possible to observe that applying a SRHT decreases the UTS of the samples for all contour conditions. From the stress-strain curves, there seems to be little variation between the tensile performance among the different contour conditions. Since slippage occurred during the test of the SB3 100-600 AB sample (indicated by an asterisk in Table 4.2), the results from SB4 are taken as a comparison point in the following paragraphs.

Consider first the AB condition. The reference 85-1200 samples exhibit the highest UTS, of approximately 455 MPa. While the UTS of 100-600 samples is similar (452 MPa), the one for 100-300 samples is considerably below the level of these (422 MPa). This behavior is not replicated after the SRHT however, as the 100-600 samples present the highest UTS (410 MPa), while that of the 85-1200 and 100-300 samples are 400 and 390 MPa, respectively. The differences between the 100-600 and 85-1200 samples are not beyond the margins of error, but the drop in UTS observed for the 100-300 samples remains consistent in both AB and SRHT conditions. Nevertheless, since both the 100-300 and 100-600 samples exhibit keyhole porosities, no clear connection can be made between the contour process parameters and this decrease in UTS.

Considering the rupture strain, it seems that the SRHT samples have a lower ductility in comparison with the AB case. In the AB condition, the fracture strain ranges from 4.4% for the 100-300 samples, to 4.7% for the 100-600 samples. After the SRHT, samples present a lower fracture strain, ranging from 3.4 % to 4% on average. The lowest fracture strain was presented by the 100-300 samples, followed by the 100-600 samples. Despite these differences, there is no clear evidence from the stress-strain curves to suggest that the process parameters alter the ductility of the samples in a significant way.

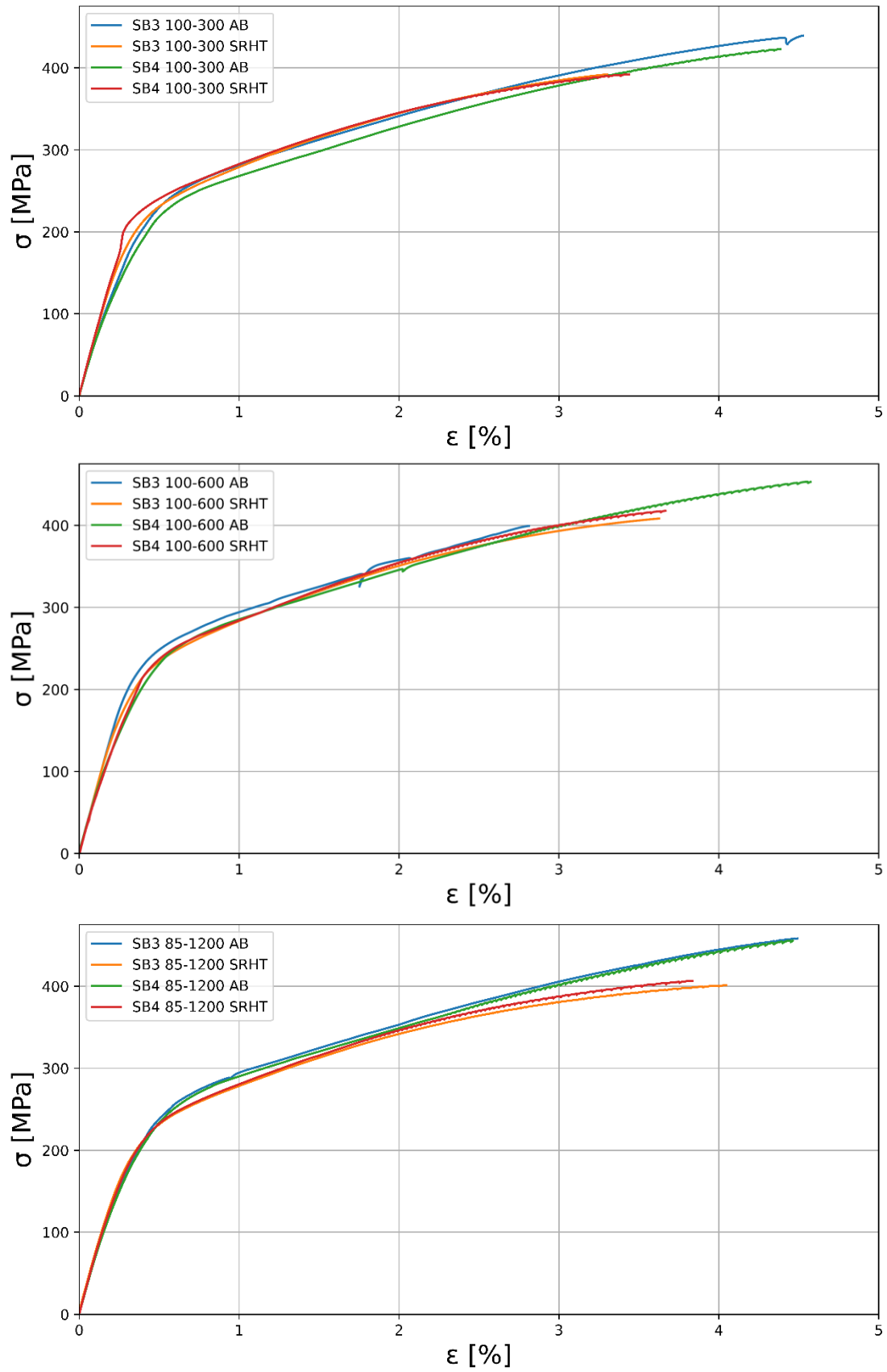


Figure 4.12. Strain-stress curves for the AB and SRHT a)100-300, b)100-600, and c)85-1200 samples.

A fractography of the SRHT samples was carried out using SEM in order to characterize the fracture behavior for the different contour conditions in more detail, and its results are illustrated in Figure 4.13. It is possible to observe from the overview of the fracture surfaces that both 100-300 and 100-600 samples exhibit a significant amount of contour porosities. This can be explained by the influence of the geometrical effect mentioned in Section 4.1, since the samples are 4 mm in diameter. The low ductility of the AlSi10Mg is evidenced by the small dimples (average size under $1\ \mu\text{m}$) present in the contour region of all samples, as seen in Figure 4.13d, e and f. However, 85-1200 samples exhibit a slightly higher number of dimples compared to the 100-300 and 100-600 samples, which may account for some of the differences in ductility observed, but it is not conclusive. From the presence of these small voids, the fracture mechanism at the contour can be described as occurring due to micro void coalescence.

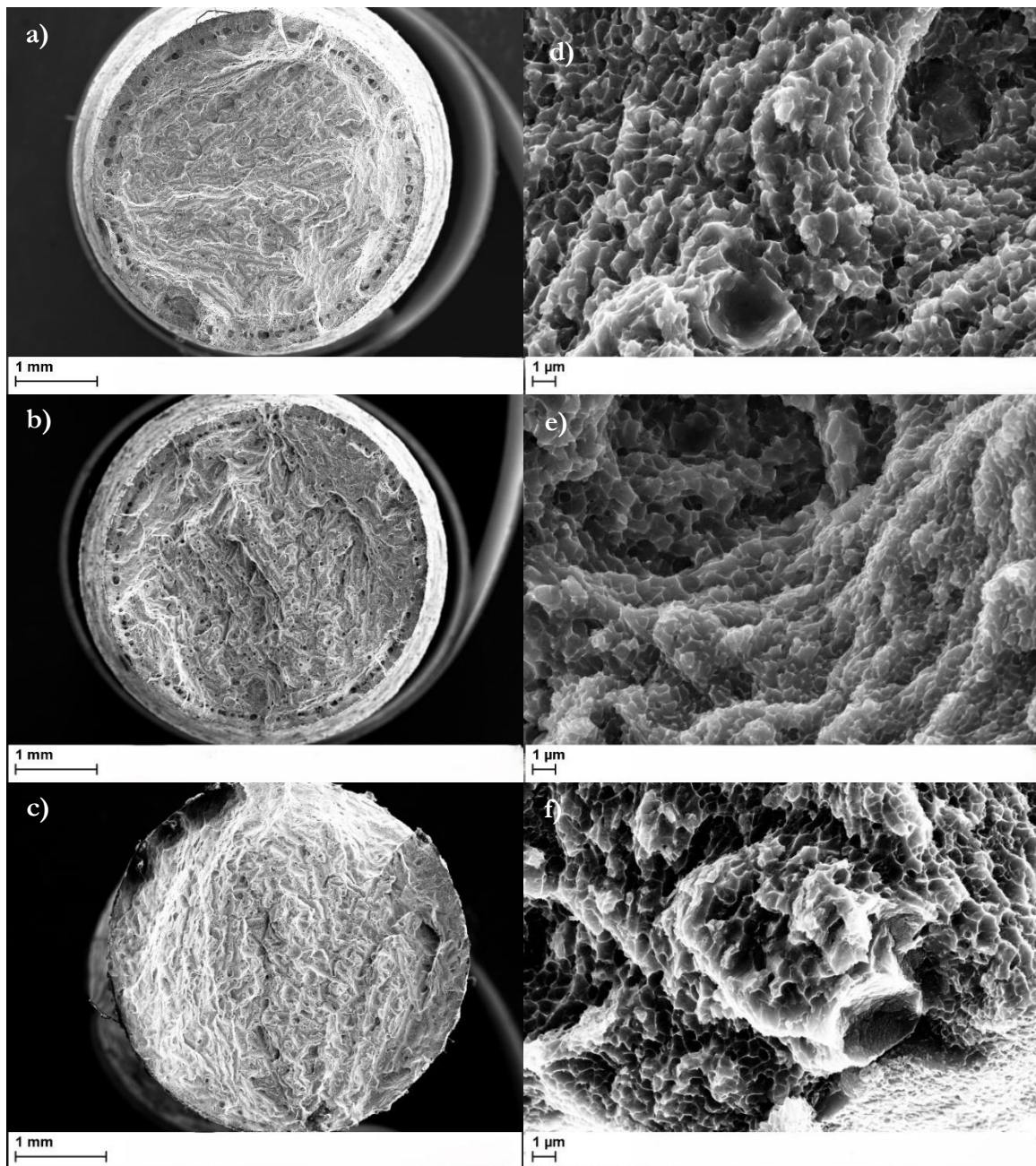


Figure 4.13. Overview and magnification of the fracture surface morphology of the SRHT samples: 100-300 a) and d), 100-600 b) and e), 85-1200 c) and f).

4.5. Fatigue Properties

The standard Wöhler curve, illustrated in Figure 4.14, present the results of the fatigue tests carried out on the SRHT samples. In addition, Figure 4.15 presents the Wöhler curves normalized to the mean yield strength of each condition. From these curves, both the 100-600 and the 100-300 samples present a better fatigue resistance than the reference 85-1200 samples. Logarithmic trendlines have been plotted for all sample groups, and they show that the 85-1200 and 100-300 samples are virtually parallel to each other. These suggests that the improvement in fatigue resistance for the 100-300 samples is uniform across all the stress levels in the tested stress range. In contrast, the trendline of the 100-600 samples presents a decline. As the ratio between the maximum stress level and the yield stress approaches 1, the 100-600 samples present a behavior comparable to that of the 100-300 samples, but as the ratio decreases, the improvement in fatigue lifetime of the 100-600 samples becomes less significant. For example, for a stress level of 220 MPa, the average increase in fatigue lifetime for the 100-300 and 100-600 samples were 84% and 64% respectively. Conversely, for a stress level of 180 MPa, the average increase in fatigue lifetime for the 100-300 samples is approximately 112%, while that of the 100-600 samples is 60%. It is important to highlight that the correlation coefficient for all three sample groups ranges from 0.82-0.86, indicating a significant scattering of data.

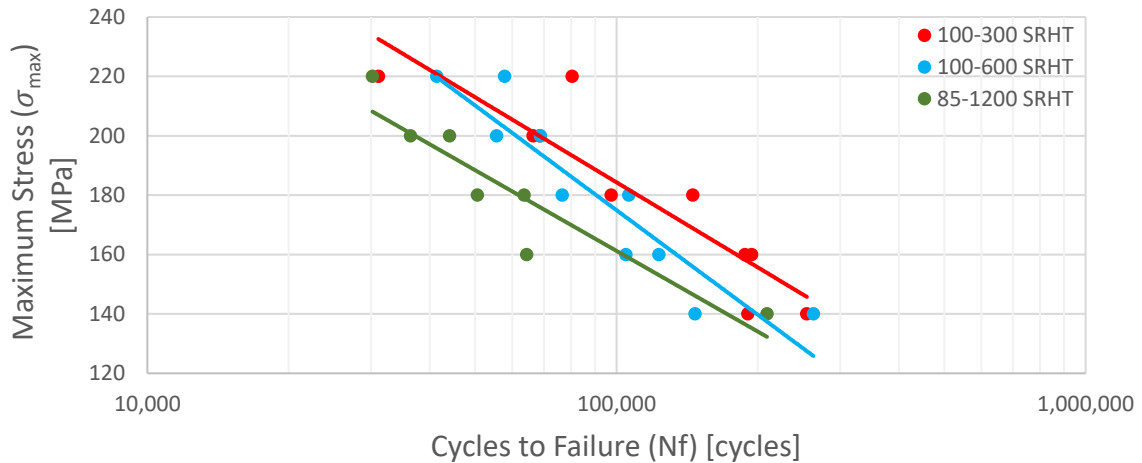


Figure 4.14. Wöhler curve for SRHT condition of the 85-1200, 100-600, and 100-300 samples

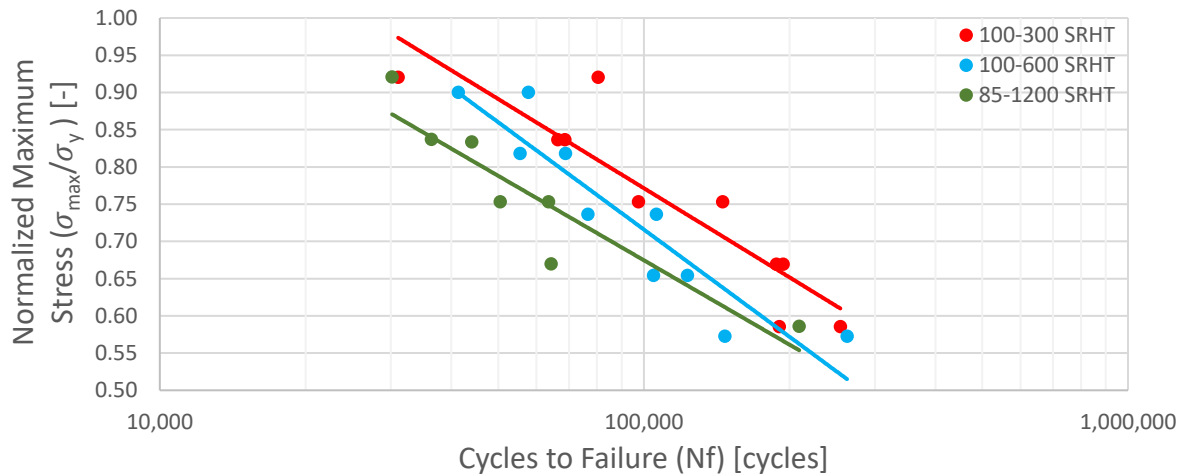


Figure 4.15. Normalized Wöhler curve for SRHT condition of the 85-1200, 100-600, and 100-300 samples.

In order to evaluate the decoupled effect of contour keyhole porosities and improved surface quality, the SRHT samples were compared with homologous sandblasted (SRHT+SB) and polished (SRHT+P) samples. The results for the 100-300 samples, illustrated in Figure 4.16, show that both polishing and sandblasting have positive effects compared to the simple heat-treated condition. The trendline for the SRHT+P samples suggests that polishing has a positive impact, which increases at lower stress levels (high-cycle fatigue conditions). On average, improvements of 59%, 78%, and 364% were observed for stress levels of 200, 180 and 160 MPa respectively, corroborating the previous observation. Between polishing and sandblasting, sandblasting was found to have the most positive effect on fatigue lifetime but produced more scattering. Although few in number, these results show that the fatigue lifetime of the 100-300 SHRT+SB samples can be increased to between 2.2 and 3.5 times that of the SRHT and SRHT+P samples.

Similar trends were observed for the 100-600 samples, as illustrated in Figure 4.17, which evidences that SRHT+P and SHRT+SB samples perform better than SRHT, particularly in lower stress levels. For a stress level of 200 MPa, the fatigue lifetime of SRHT+P is 3 times that of SRHT, while for lower stress levels the lifetime difference is an order of magnitude higher, reaching beyond 6 million cycles. Furthermore, for a stress level of 160 MPa, the 100-600 SRHT+P sample did not break after 10 million cycles. One of the SHRT+SB samples performed better than the SRHT at 200 MPa, but a significant scatter, encompassing almost 2 orders of magnitude, is present.

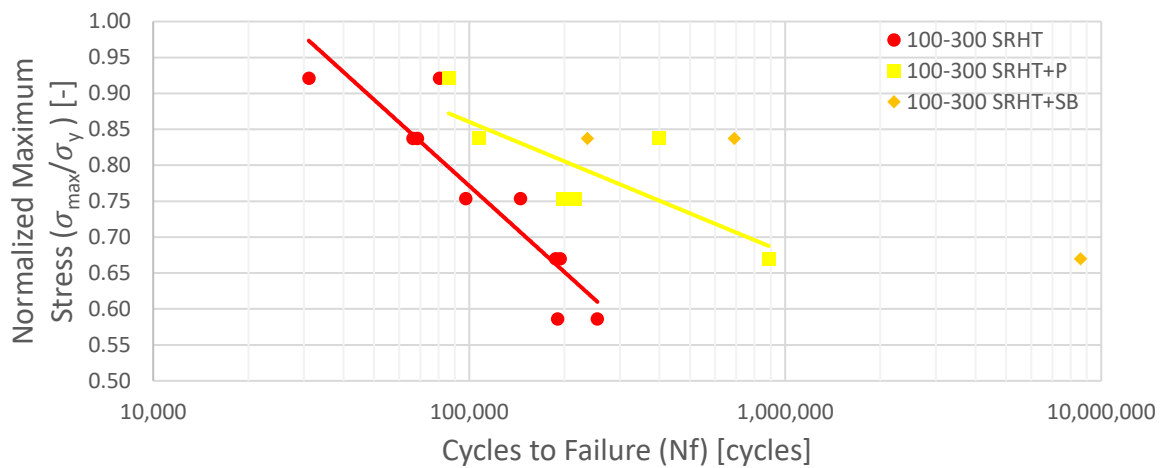


Figure 4.16. Normalized Wöhler curve for the 100-300 samples in SRHT, SRHT+P, and SRHT+SB conditions.

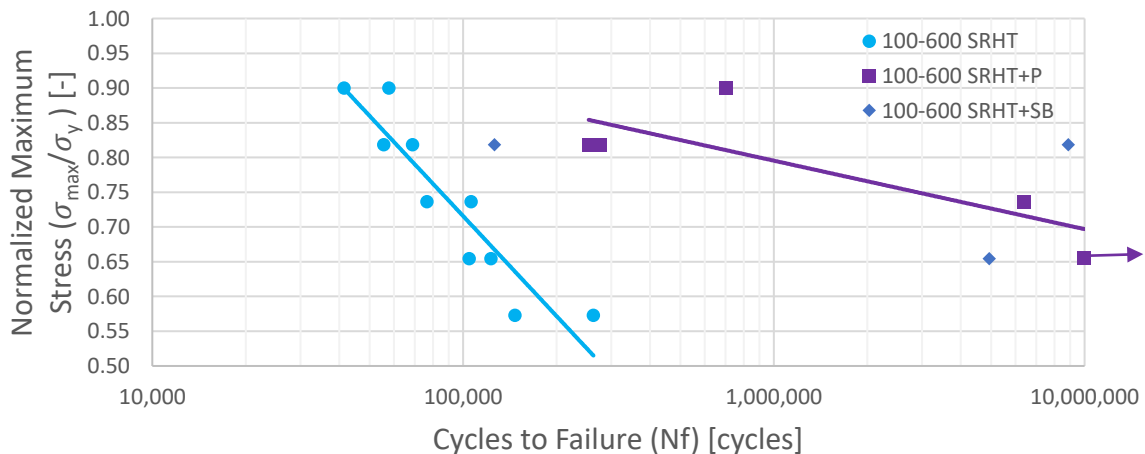


Figure 4.17. Normalized Wöhler curve for the 100-600 samples in SRHT, SRHT+P, and SRHT+SB conditions.

A fracture surface analysis was carried out on the fatigue specimens to gain a deeper understanding of the fatigue mechanisms involved. Figure 4.18 provides an overview of the fracture surface of a 100-300 SRHT sample tested at 200 MPa, which exhibits the major characteristics observed for all contour conditions. It is possible to identify three distinct regions within the surface:

- Region 1: Corresponds to the crack nucleation site and slow crack growth region. In most specimens, crack nucleated at the surface, most likely due to the surface roughness. In this 100-300 SRHT sample, as seen in Figure 4.19a, it is possible to observe that the crack nucleated in the site where a keyhole was located close to the surface. The slow crack growth region is characterized by a relatively flat surface marked by river marks and ridges, as shown in Figure 4.19b. This indicates that throughout this region, the crack advanced through pseudo-cleavage.
- Region 2: Characterized by striations perpendicular to the crack propagation direction, as seen in Figure 4.19c which indicate a stable crack growth. An analysis at a higher magnification (not shown) revealed that in this region, the tearing was semi-ductile in nature: small dimples were observed, but the striations appear to be separated by jagged edges, indicating brittle-like behavior. This is consistent with the previous results, which shows that this alloy has a low ductility.
- Region 3: Characterized by a cleaved surface, reminiscent of those observed in static fracture. Figure 4.19d illustrates the transition from Region 2 to Region 3 observed in the contour region. Micro porosities are observed at the boundary between contour and bulk, as well as the keyholes in the center of the contour, as noted in Figure 4.5.

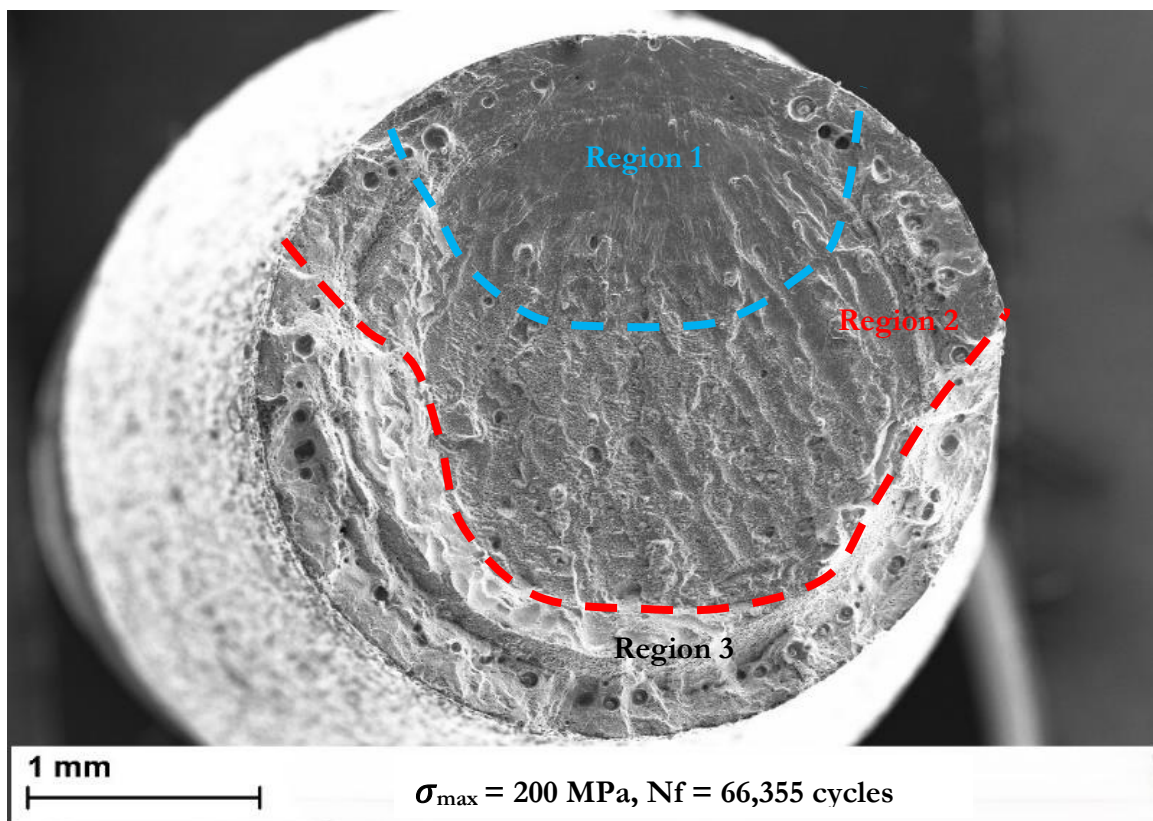


Figure 4.18. Overview of the fracture surface of a 100-300 SRHT sample.

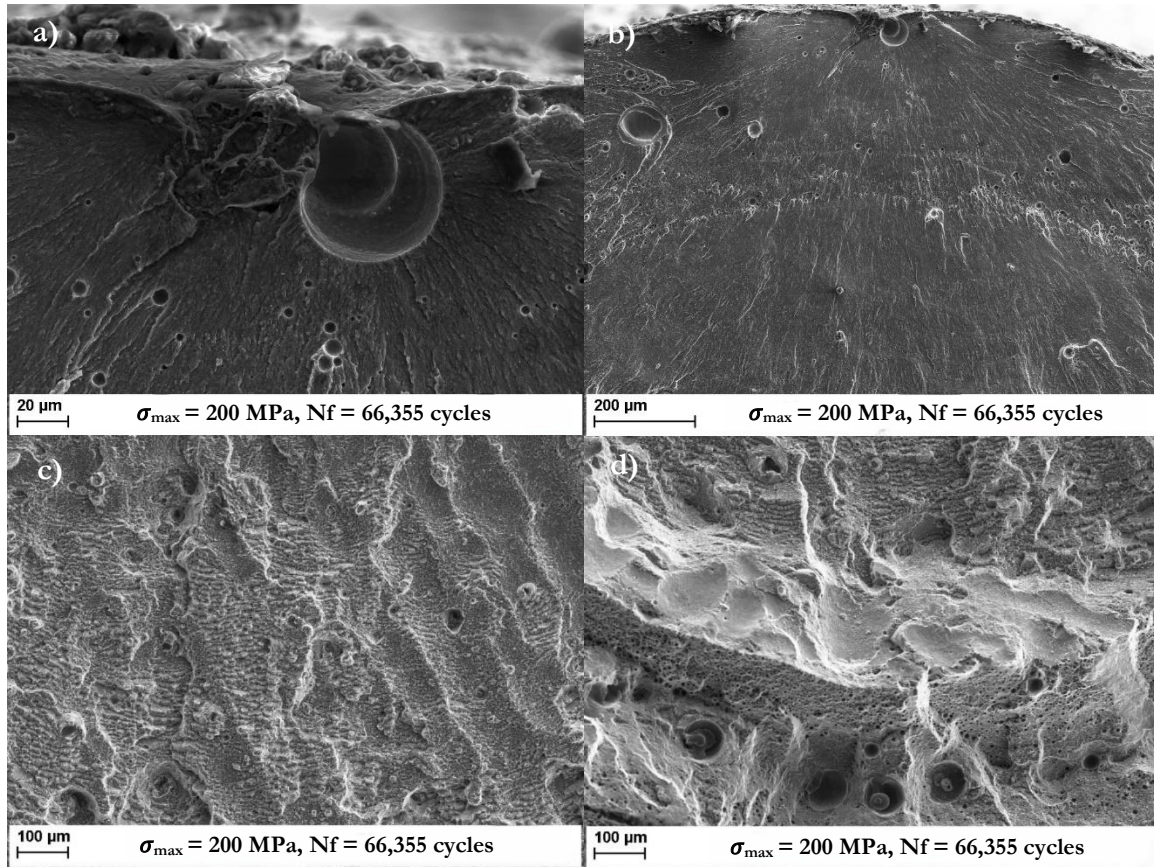


Figure 4.19. a) Crack nucleation site, b) slow crack growth, c) stable crack growth, and d) fast fracture regions in a 100-300 SRHT sample.

Even though there were similarities in terms of the fracture behaviors observed among all sample groups, there were differences in the crack nucleation sites and size of each region at different stress levels. Figure 4.20 shows the crack nucleation sites in all sample groups for a stress of 200 MPa. Notice that in all cases, the initiating defect is in the surface and there are visible ratcheting marks in the vicinity of the crack initiation site. However, the size of the initiating defect for 100-300 and 100-600 samples is much smaller than that of 85-1200 samples. Furthermore, it is possible to observe that the initiating crack zone for 85-1200 samples has a larger number of ratcheting marks, indicating the presence of more microcracks generated from the surface. These results are consistent with the roughness measurements, since the 85-1200 samples have a higher surface roughness than both the 100-300 and 100-600 samples, which could explain both the propensity for crack nucleation as well as the size of the initiating defect.

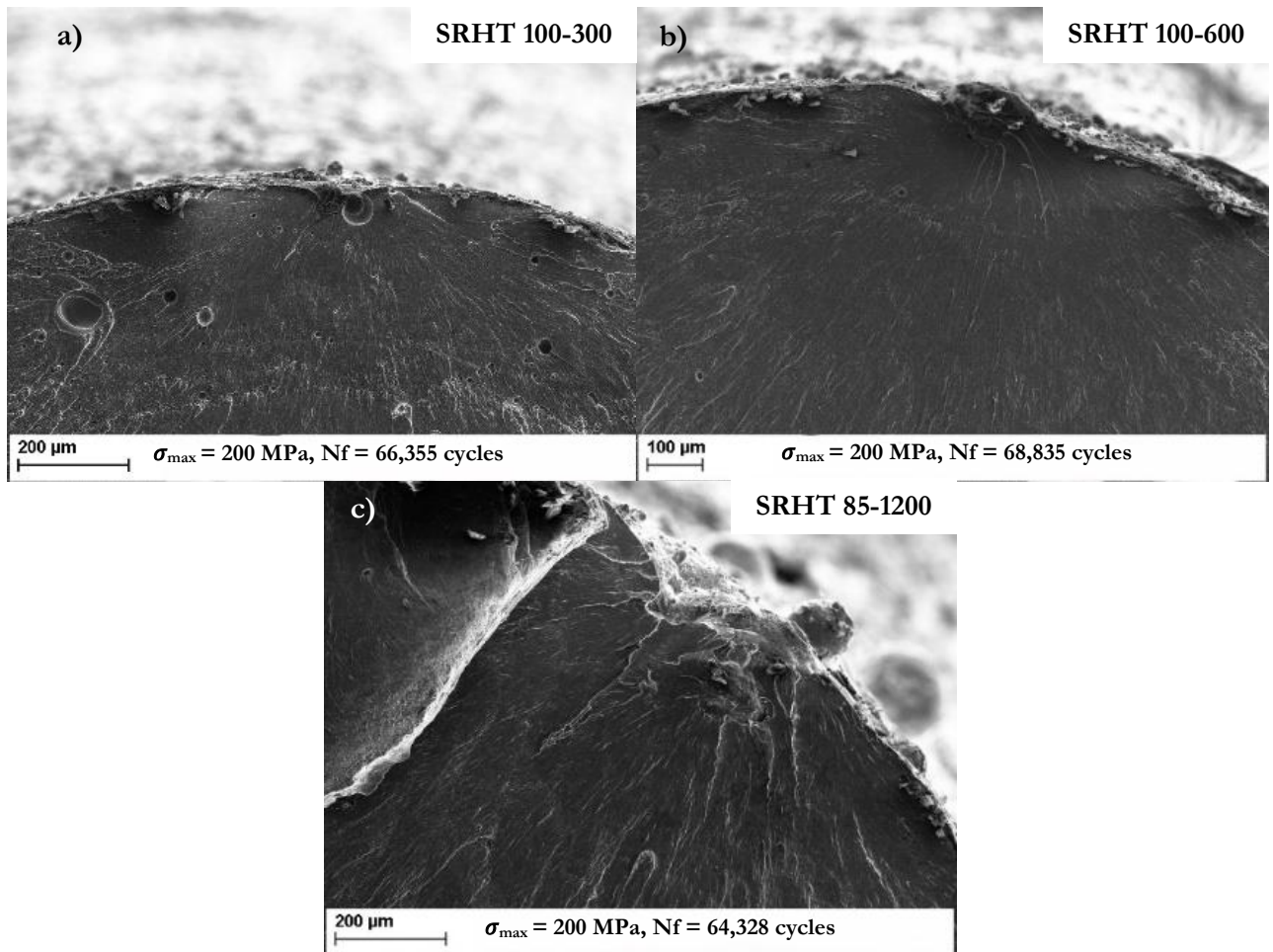


Figure 4.20. Crack nucleation sites in SRHT a) 100-300, b) 100-600 and c) 85-1200 samples.

Figure 4.21 depicts the crack growth regions in all sample groups for a high (220 MPa) and a low (140 MPa) stress level. In all sample groups, when the specimens are subjected to a high stress level, the crack nucleation and slow growth region (blue dashed line) is smaller than the stable crack growth region (red dashed line). In contrast, when the specimen is subjected to a low stress level, the slow growth region is significantly larger, and consequently, the stable crack growth region is smaller.

Differences can be found in the way the fracture regions are distributed among the sample groups. For 100-300 samples, as evidenced in Figure 4.21a and b, the stable crack growth region is mostly limited to the bulk, and the fast fracture is observed in the contour. A similar behavior is observed in the 100-600 sample at 220 MPa, but not at 140 MPa. At the lower stress level, the stable crack growth region loses its hemispherical profile matching the contour, as evidenced in Figure 4.21c and d. In 85-1200 samples, the stable crack growth region is much smaller than in the 100-300 and 100-600 samples at 220 MPa, and does not seem to match with the contour, as seen in Figure 4.21e and f.

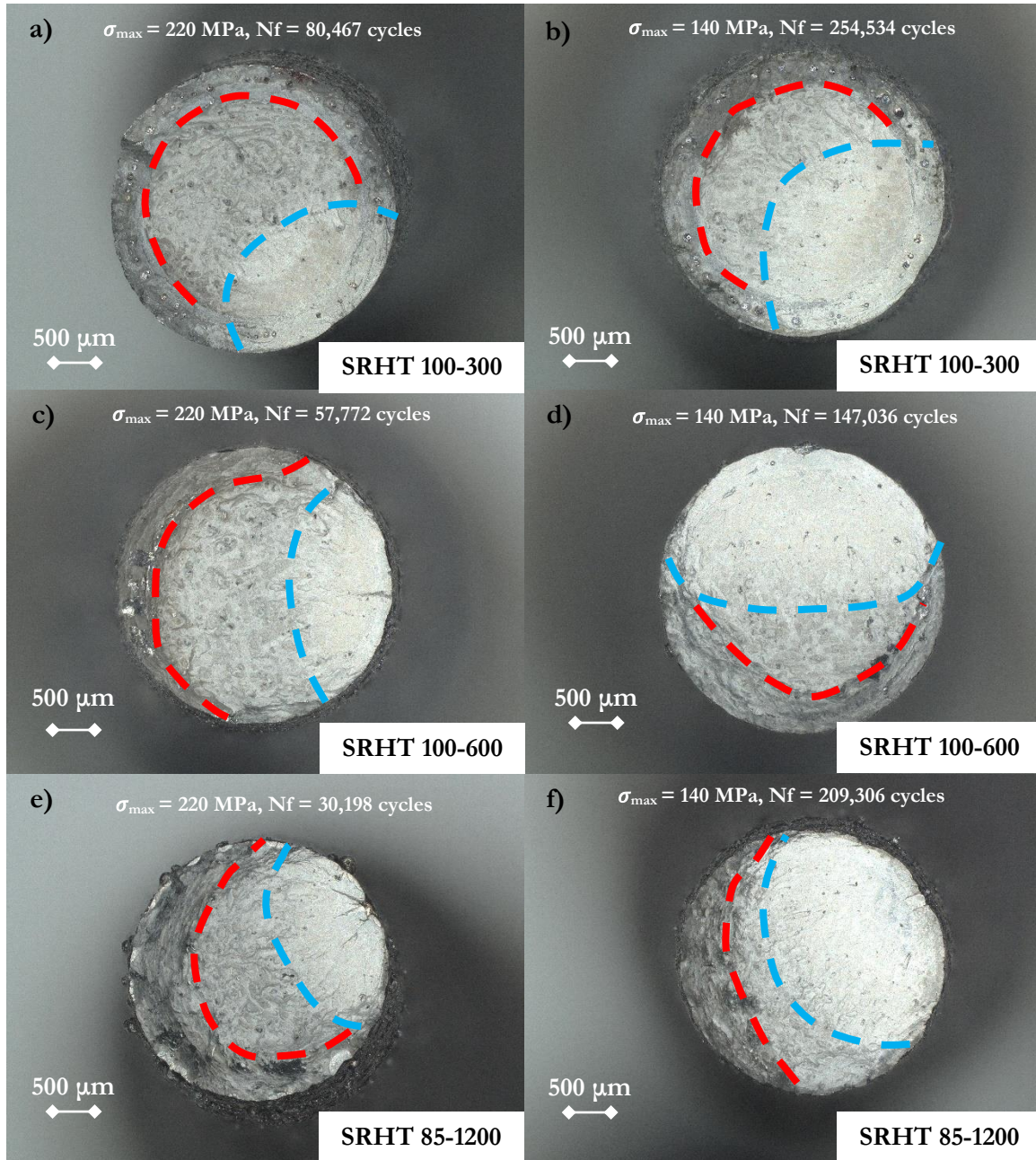


Figure 4.21. Fracture at 220 and 140 MPa for SRHT 100-300 (a and b), 100-600 (c and d), and 85-1200 (e and f) samples.

To compare the fatigue behavior of the SRHT, SRHT+P, and SRHT+SB samples, a fractography was carried out on the SRHT+P and SRHT+SB 100-300 specimens with shortest lifetime at 200 MPa, shown in Figure 4.22. In general terms, the fracture behavior in both cases is similar to that of the SRHT sample observed in Figure 4.18, as the three zones exhibiting different crack propagation characteristics are visible. Additionally, for all sample groups, the fracture initiated at the surface despite the improved surface condition, particularly in the SRHT+P samples. The slow crack growth region (blue) in SRHT and SRHT+P samples is of comparable size and similar in topology, while in the SRHT+SB sample it is smaller, although it presents more prominent river marks and ridges.

The most significant differences between the SRHT and SRHT+P compared to the SRHT+SB sample appear mainly in the stable crack growth region. Figure 4.19c and Figure 4.22c show that both the SRHT and SRHT+P samples exhibit mostly ductile tearing in the stable crack growth region. In contrast, Figure 4.22d shows that the SRHT+SB sample exhibits a stable crack growth region marked by smoother surfaces and less striations. An analysis at higher magnification (not shown) in the regions where striations are present revealed the presence of dimples, which indicate ductile tearing. The smooth regions, marked by curved topology, may correspond to the border of the melt pools.

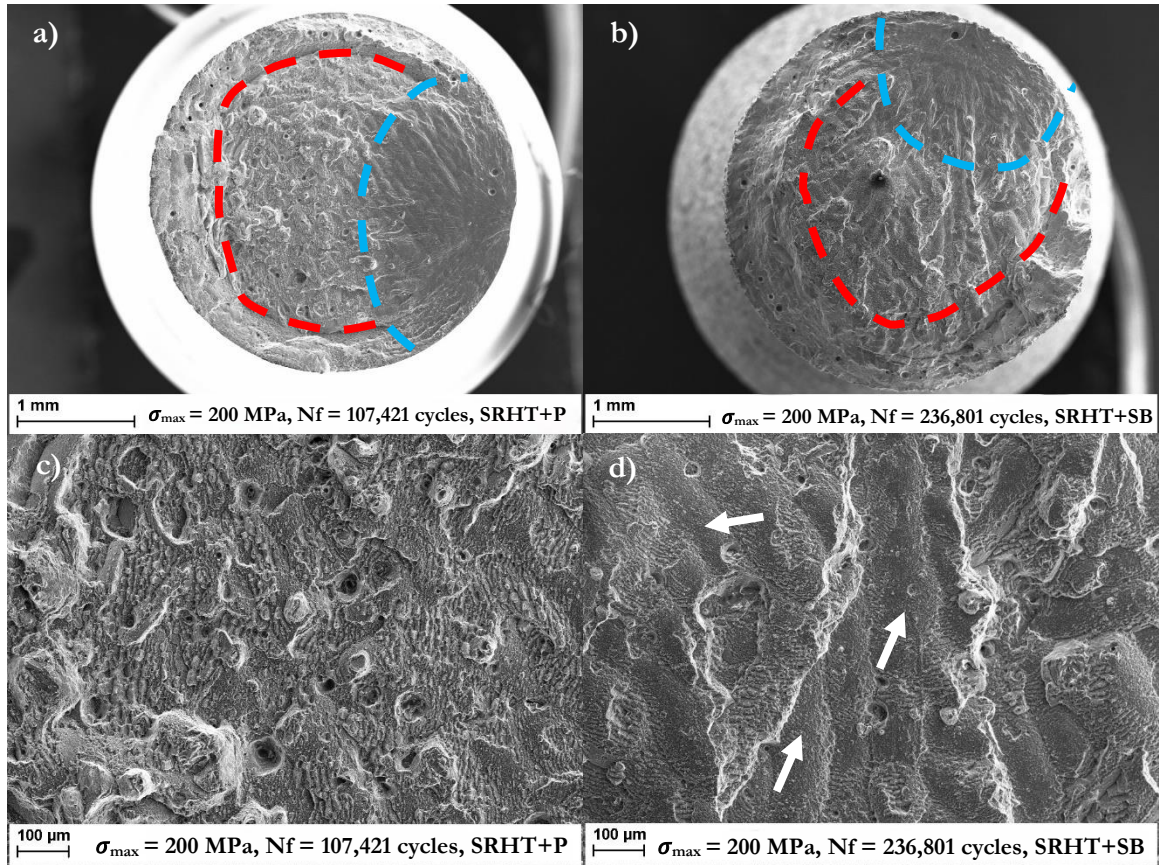


Figure 4.22. Overview and stable crack growth region in a polished (a and c) and sand blasted (b and d) 100-300 samples.

Despite the limited number of points, it is possible to observe a higher scattering in the fatigue results of SRHT+SB than in the SRHT and SRHT+P samples, particularly in the case of the 100-600 samples. To assess the differences in each case, the fractography results of each SRHT+SB 100-300 and 100-600 sample at 200 MPa are presented in Figure 4.23.

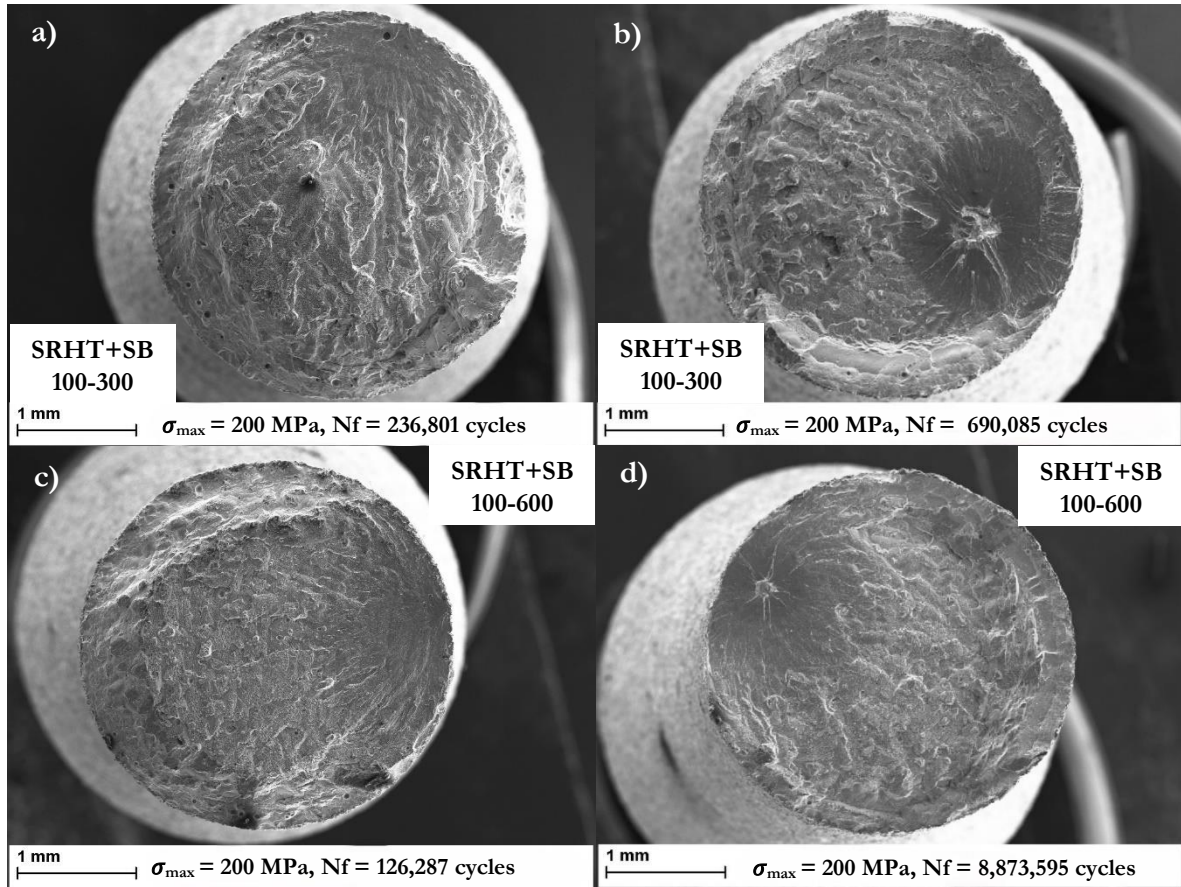


Figure 4.23. Fractography of the shortest (a) and longest (b) SRHT+SB 100-300 sample and the shortest (c) and longest (d) lived SRHT+SB 100-600 sample at 200 MPa.

The fracture surfaces exhibit the same crack growth regions that were observed in previous samples, but the most notable difference is the crack nucleation site. In all of the SRHT 100-300 and 100-600 samples, the crack started at or very near the surface at all stress levels. The same behavior can be observed for the shortest-lived 100-300 and 100-600 samples (Figure 4.23a and c), but not in the longest-lived samples (Figure 4.23b and d). In the case of the longest-lived 100-300 and 100-600 samples, cracks nucleated from internal defects rather than the surface. In the case of the 100-300 sample, the defect is about 300-400 μm across, and is located in the bulk region. In contrast, the 100-600 sample presents a smaller defect of 100-200 μm across and is located closer to the contour edge.

The differences in crack nucleation sites for the SRHT+SB 100-300 and 100-600 sample groups at the same stress level (200 MPa) highlight the impact of the sensitivity of fatigue samples to sandblasting. Sandblasting is able to minimize the impact of roughness inherent to fatigue and enhance performance by forcing the crack to nucleate in the bulk rather than the surface. However, minor variations in the sandblasting protocol may create preferential crack nucleation sites in the surface, which will decrease the fatigue life. This is evidenced by the fact that in the case of the 100-300 sample, the ratio between the fatigue life of longest and shortest-lived samples is 2.9, while for the 100-600 samples it is 70.2. How sandblasting is able to minimize the impact of surface roughness is further discussed in Section 5.4.

5. Discussion and Analysis

The following section aims to provide an analysis of the experimental results presented in Section 4 and a comparison to relevant experimental results from the literature. The discussion starts by exploring the relationship between the process parameters and geometry to the formation mechanisms behind surface roughness and porosities. This is followed by an analysis of the mechanical performance in tensile and fatigue tests, and how it relates to microstructure, roughness, and porosity. A concerted effort has been made to put forth explanations which can provide insight into the phenomena observed during the experimental work.

5.1. Impact of Process Parameters on Keyhole Formation and Surface Roughness

The impact of varying the process parameters (in particular, the scanning speed for a given laser power) has been presented in Sections 4.1 and 4.2. For a fixed geometry, it is possible to improve the lateral surface quality of the samples by increasing the TED from 193 to 1368 J/m. This was achieved by decreasing the contour scanning speed from 1200 to 200 mm/s. The result was a decreased surface roughness, coupled with the gradual emergence of keyhole porosities close to the surface. As mentioned in Section 2.1.4, all other parameters being equal, the laser power and scanning speed will determine the melting mode during L-PBF. As a result, a higher contour TED will create the necessary conditions to create keyholing, while a lower contour TED will give rise to a transitional or conduction mode.

To evaluate the effect of the contour parameters on the melt pool, cross-sectional views (parallel to the building direction) of the core section of 100-300 and 100-600 AB half samples are presented in Figure 5.1. It is possible to observe that the dimensions of the melt pool in the 100-300 samples are larger than those in the 100-600, particularly in terms of the width of the contour and total height of the melt pool. This behavior was analyzed for all the contour conditions evaluated in the OB and are illustrated in Figure 5.2 and Figure 5.3

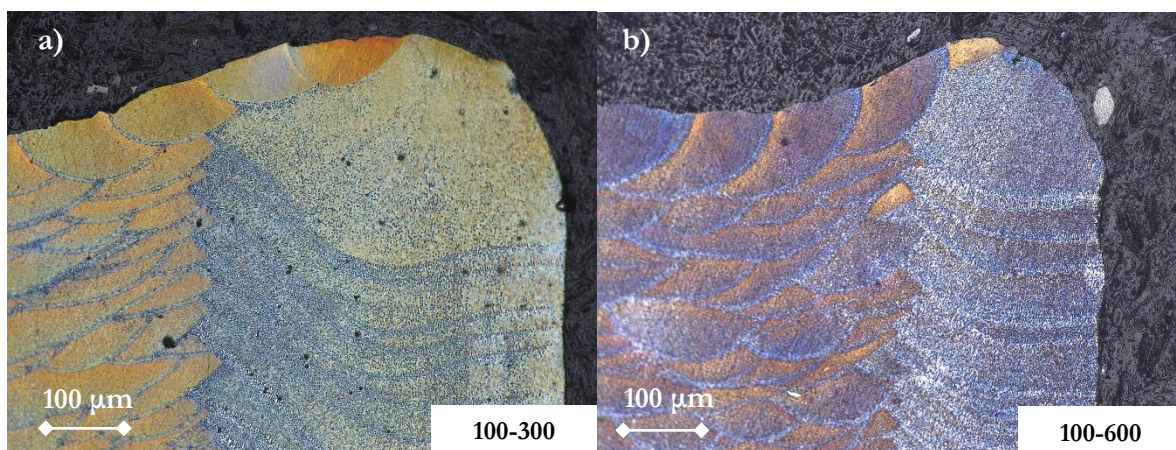


Figure 5.1. Cross-sectional view of the contour melt pools parallel to the building direction at the core of a) 100-300 and b) 100-600 half samples.

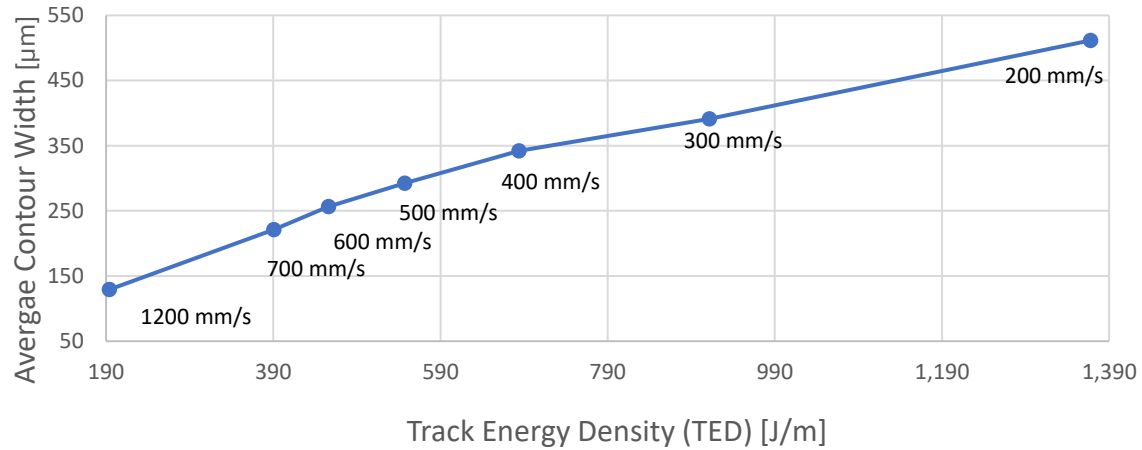


Figure 5.2. Contour width as a function of TED.

The relationship between melt pool and scan track dimensions with the process parameters has been extensively studied for a variety of metallic alloys. Experimental results show that increasing the TED, either by decreasing the scanning speed or increasing the laser power, tends to create wider and deeper melt pools [53]. A comprehensive investigation into the L-PBF of AlSi10Mg [54] showed that for a fixed laser power, increasing the scanning speed from 200 to 1000 mm/s (TED range from 1250 to 250 J/m) caused a quasi-exponential decay in the melt pool's depth, width and depth-to-width ratio. Beyond this speed range, further increases in the scanning speed produced no significant differences. These results are consistent with the results presented in this investigation and suggest that the selected process windows at 300 mm/s and 600 mm/s are situated within the range in which the melt pool is highly sensitive to morphological changes.

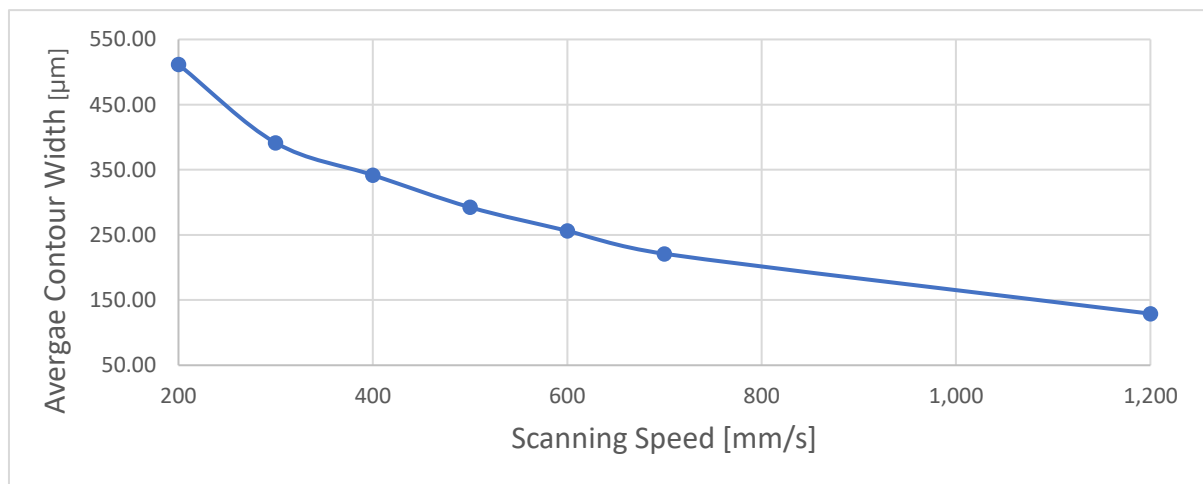


Figure 5.3. Contour width as a function of scanning speed.

The presence of contour keyholes as observed in Figure 4.5 for the 100-300 samples suggests that by operating at 300 mm/s (TED = 912 J/m), vapor depression instabilities are likely to arise. The high depth-to-width ratio evidences the melt pool instabilities, and trapped gas bubbles that are unable to escape the melt pool are likely to be created. The relationship between a higher TED and a higher probability of keyholes is corroborated by the differences observed in the depth of the melt pools at the head and core of the 100-300 samples, seen in Figure 5.5. The melt pools in the head a higher depth-to-width ratio, leading to more and larger keyholes in the head compared to the core. The reasons for the differences in the melt pools at the core and head of the same

sample are linked to the localized heat input and dissipation as a result of the geometry effect. This will be further discussed in Section 5.2.

It is important to highlight that the normal keyhole formation mechanism might not be able to explain the microporosities observed close to the contour edges seen in Figure 4.5. These microporosities are more numerous in the 100-300 samples than in the 100-600 samples, and their presence in both suggests a possible link to the TED. One possible explanation for this is the formation of hollow and semi-hollow spattering particles that occurs at high TED levels. An investigation into the relationship between energy intensity and spattering in AlSi10Mg [55] poses that metal evaporation will cause molten droplets to be ejected from the melt pool, which are mainly governed by the gas-liquid interfacial tension. These droplets may trap shielding gas, forming small pores after solidification, as illustrated in Figure 5.4. These results are consistent with the pores observed at the edges of the contour, as the investigation reports pores varying from 4 to 20 μm in size, which match the size of the pores observed in this work.

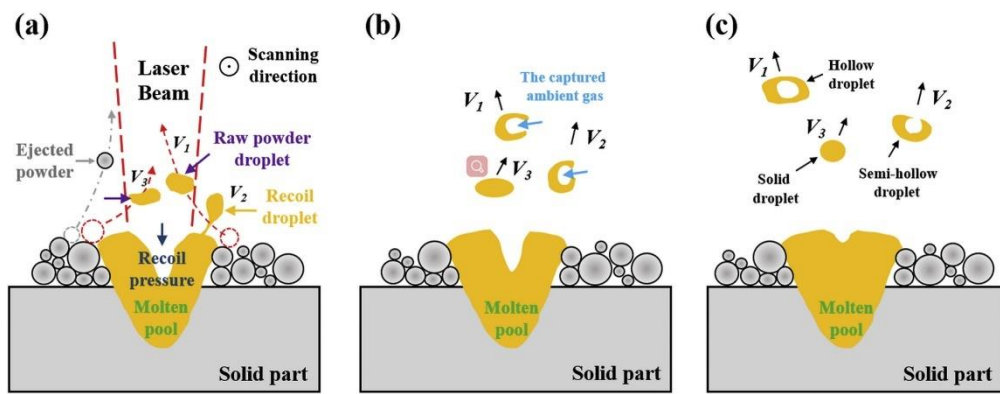


Figure 5.4. Formation mechanism of the microporosities observed at the edges of the contour scan track. [55]

Another interesting impact of a high TED is the overlap between successive tracks. Sufficient remelting of previous tracks is necessary in order to ensure good adhesion and avoid harmful lack of fusion porosities. However, as seen in Figure 5.5b, an excessive overlap between the bottom of the melt pools might cause large keyholes at the center of melt pools from successive layers to coalesce. This coalescence changes the shape and size of the keyholes, since the porosities lose their sphericity and exhibit an irregular form with sharp edges, which could become detrimental for mechanical applications.

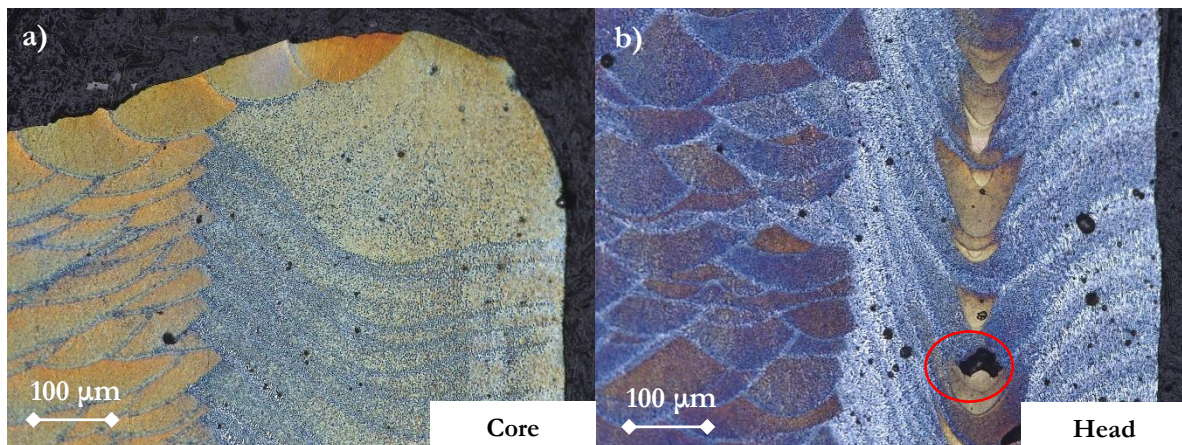


Figure 5.5. Differences in depth-to-width ratio and keyhole coalescence in the contour at the a) core and b) head of a 100-300 half sample.

The geometrical differences in the melt pools of 100-300 and 100-600 samples, which influence the creation of keyholes, also account for the differences in roughness, particularly with respect to the 85-1200 samples. A comprehensive investigation into the relationship between stability and melt pool geometry on aluminum alloys [56] revealed that different stability process windows will lead to different track morphologies. It is possible to create thick and continuous tracks characterized by a greater depth-to-width ratio, most likely corresponding to the onset of keyholing. According to a previous investigation into the improvement of surface quality by the optimization of process parameters [10], increasing the TED causes the radius of curvature of the melt pool to increase, creating an almost flat side profile. This flat side profile was found to improve the lateral surface roughness. This is consistent with the side profile of the 100-300 and 100-600 samples seen in Figure 5.1. In both cases, the TED used to manufacture the samples causes the side profiles to be almost flat, which explains the low roughness of both building parameters.

Coupled with the geometrical changes induced in the melt pools by the TED level, the improved surface roughness can also be attributed to the thermophysical differences at the contour due to the energy input. Experimental results on AlSi10Mg [54] have reported that, for a fixed laser power, increasing the scanning speed will increase the associated cooling rate. The experimental results match the analytical behavior expected from Rosenthal's 2D equation, in which the cooling rate is inversely proportional to the TED. Since 100-300 and 100-600 samples will present a lower cooling rate compared to 85-1200 samples, the contour will remain at higher temperature for a longer time period of time. This high temperature latency will favor wettability and the reduction of viscosity and surface tension, which will inhibit the adherence of particles. This could explain why increasing the TED tends to decrease the peak more than the valley roughness, as observed in Figure 4.10.

5.2. Impact of Geometry on Keyhole Formation and Surface Roughness

The results of the selection and optimization process described in Section 4.1 show that the geometry of the built specimen will have an impact on its thermal history, microstructure and mechanical properties. Microscopic observations at the head and core (Figure 10.1) and the parametric study at different scanning speeds (Figure 4.2 and Figure 4.3) correlate the TED with localized heat accumulation in geometries with a larger cross sectional area. This is supported by the fact that for similar process parameters (same TED), there were significant differences in the porosity and roughness observed at the core and head of each sample. Furthermore, from the parametric study of samples with 2 and 3 mm core diameters, different porosity and surface roughness levels were observed. From these results, it seems that the geometrical effect plays an important role on the heat input and dissipation mechanisms and will significantly impact the roughness and porosity.

To assess the impact of the cross-sectional area on the porosity distribution, an in-depth analysis of the CT scans of the 100-300 and 100-600 samples with a 3 mm core diameter was carried out using Avizo. The schematic representation of this analysis is presented in Figure 5.6. The CT scan of each sample was subdivided into sub-volumes with a thickness of 1.5 mm. In each sub-volume, segmentation and thresholding were used to define the contour and bulk regions, and porosity levels in each of these regions calculated. For both the 100-300 and 100-600 samples, the analysis of the contour porosities was limited to a depth of 300 μm from the surface. As a result, the porosity level in each sub-volume is associated to the mean diameter of the corresponding sub-volume. Characterization of the microporosities formed from semi-hollow ejected particles was out of the scope of this analysis due to the voxel size used in the CT scan (6 μm).

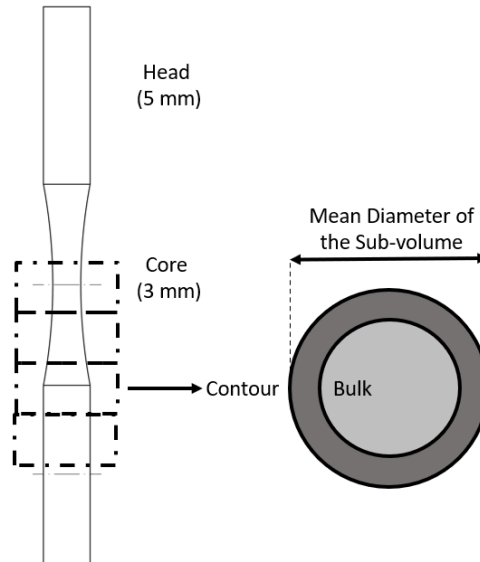


Figure 5.6. Schematic representation of the porosity analysis in the CT scans of the 100-300 and 100-600 samples.
Drawing not to scale.

The results of the porosity analysis are presented in Figure 5.7. In both the 100-300 and 100-600 samples, the porosity distribution in the contour and bulk as a function of the mean diameter presents two distinct behaviors. The contour porosity increases with the diameter, reaching a maximum diameter of approximately 4.25 mm, followed by a slight decrease and stabilization at 5 mm. The contour porosity level reached by the 100-300 sample is greater than that reached by the 100-600 sample, as expected due to the higher contour TED in the 100-300 samples. In contrast, the bulk porosity increases continually as the mean diameter increases. Notice that the bulk porosity at a mean diameter of 5 mm (corresponding to the head of the sample) is almost 4 times higher for the 100-600 samples than for the 100-300 samples. This confirms the results observed in Figure 4.7.

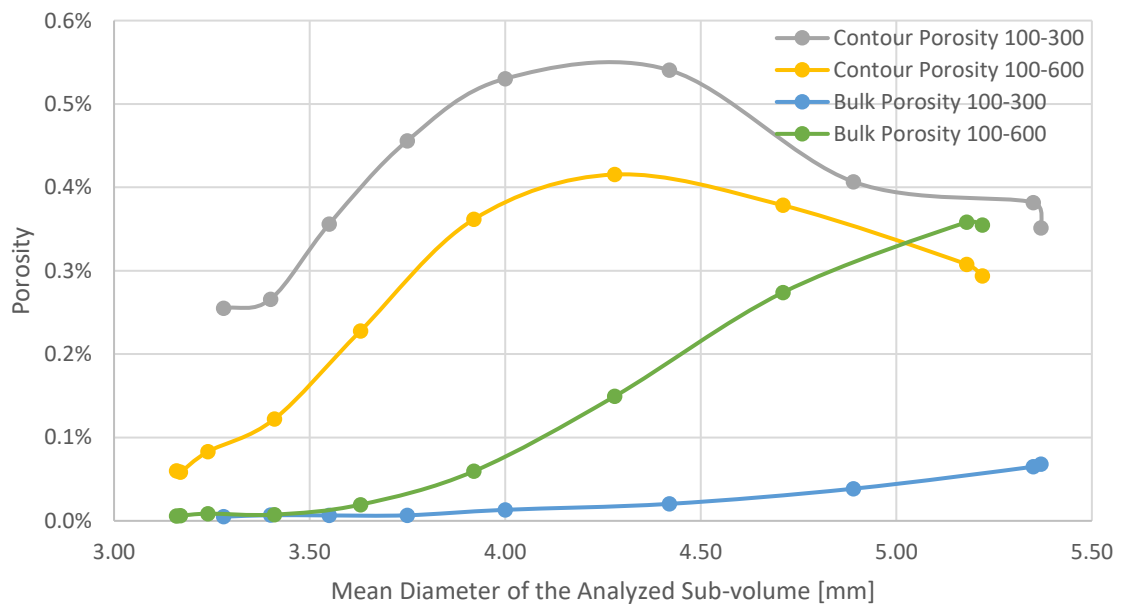


Figure 5.7. Porosity distribution as a function of diameter in 100-300 and 100-600 samples.

To provide a possible explanation for this behavior, it is necessary to look into the heat transfer mechanism at play in the core and head of the samples for different specimen geometries. As discussed in Section 2.1.4, the dominant melting mode will depend on the process parameters (in this case, laser power and scanning speed), which will impact the shape and size of the melt pool. Conduction and keyhole modes, illustrated in Figure 2.6, are related to low and high energy inputs, respectively. Considering only the TED, a preliminary assumption would be that, regardless of the cross-sectional diameter, the melting mode and melt pool behavior should be the same. However, the experimental results from the parametric study (presented in Figure 4.2 and Figure 4.3), showcase that, for the same contour parameters, keyholes will be observed for different scanning speeds in 3 mm samples compared to 2 mm samples. Moreover, within the same sample, Figure 5.5 and Figure 5.7 showcase how the melt pool shape and porosity distribution change at the head and core regions of the same 100-300 half sample.

Since the TED alone is unable to explain the differences in the thermal history in the head and core for the 3 and 2 mm core diameter samples, an alternative approach consists of evaluating the heat input and heat dissipation associated to each geometry. The heat input is related to the total amount of power that the laser injects into each powder layer, while the heat loss occurs due to the contact with the surrounding unfused metallic powder and with the build plate. Consider first the heat input into each sample. The energy input in the core and the bulk for different cross-sectional diameters can be determined using the number and length of tracks in the bulk, the contour track length, and the TED. Table 5.1 presents the heat input in the head and core of 2 and 3 mm samples. The contour process parameters are 100-300 and 100-600, and the bulk process parameters are 85-1200 (see Table 4.1).

Core Diameter	Contour				Bulk	
	Core		Head (5 mm)		Core	Head (5 mm)
	100-300	100-600	100-300	100-600	85-1200	
2 mm	5.56	2.78	14.15	7.08	5.71	37.16
3 mm	8.42	4.21			13.15	

Table 5.1. Heat input, in Joules [J] at the head and core of 2 mm and 3 mm samples in the contour and bulk regions.

Two different observations can be made regarding the total heat input:

- Consider the heat input at the core. The total contour heat input in the core in 3 mm samples is 1.5 times higher than in 2 mm samples for both contour conditions. Furthermore, the contour heat input in the head is 1.75 higher than in the core for the 3 mm samples, and 2.5 times higher for the 2 mm samples.
- Consider the heat input at the bulk. The total bulk heat input in the core is 2.3 times higher in the 3 mm samples compared to the 2 mm samples. Moreover, the bulk heat input is 2.8 time higher in the head than in the core for the 3 mm samples, and over 6.5 times higher for the 2 mm samples.

These observations explain the behavior observed in the parametric study for the 2 and 3 mm core diameter samples. In the core, 3 mm samples have a higher total heat input than the 2 mm samples for the same TED. The lower heat input leads to the absence of keyholes and higher surface roughness in the 2 mm samples compared to the 3 mm samples. Furthermore, for any given sample, the total heat input in the head is considerably larger than in the core. To study the effect of the heat input on the melt pool morphology, a view parallel to the building direction at the head and core sections of a 100-600 sample are presented in Figure 5.8.

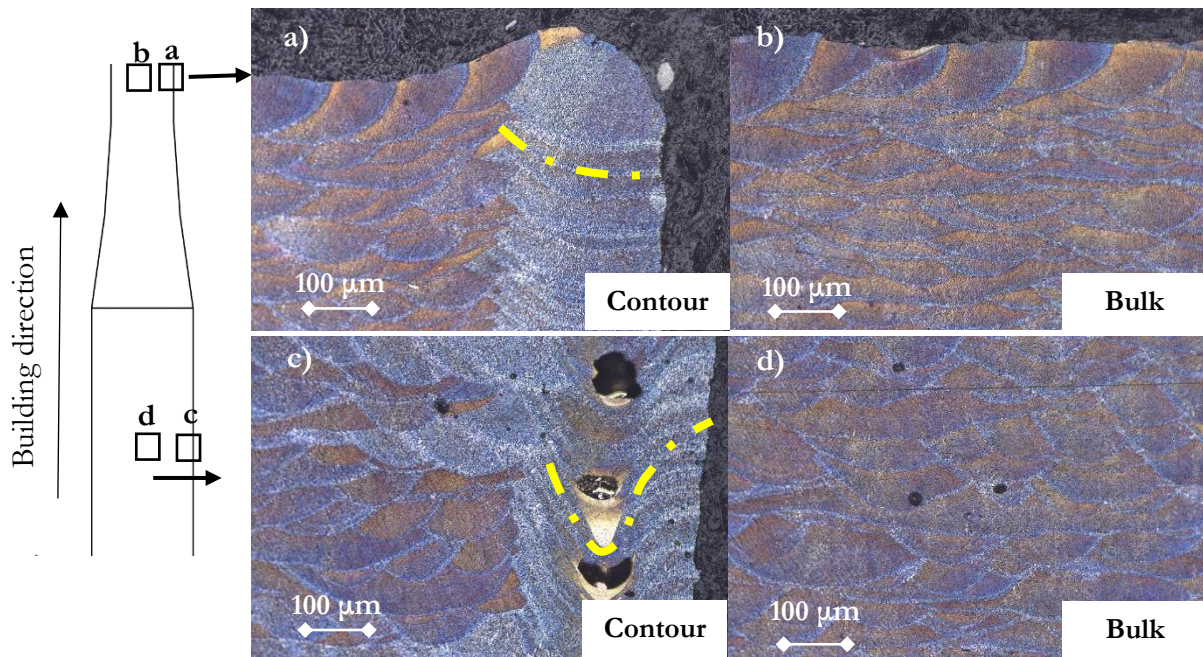


Figure 5.8. Cross-sectional view parallel to the building direction of melt pools in a 100-600 half sample. A) Contour at the core, b) bulk at the core, c) contour at the head, d) bulk at the head.

Consider the melt pools at the contour in the core and head sections of the sample. In the core (Figure 5.8a), the depth of the melt pool is similar to its width, and no keyholes are visible. In contrast, the contour at the head (Figure 5.8b) is characterized by melt pools which have a similar width to those in the core, but a greater depth. Additionally, keyholes are visible at the bottom of the melt pools at the head. This difference in melt pool depth is not restricted to the contour of the specimen and is also visible in the bulk. By comparing the bulk zone in core (Figure 5.8c) and head (Figure 5.8d), it is evident that the melt pools in the core are shallower compared to those in the head. Moreover, it is possible to observe a keyholes at the bottom of some of the melt pools. These results are consistent with the expected behavior given the heat input. In general, both the bulk and contour regions of the head core have a higher heat input than the bulk and contour in the core. The higher heat input leads to the deeper melt pools with a flat side profile, causing lower roughness as well as keyholes.

While the heat input can be determined based on the geometry and process parameters, the heat dissipation mechanism is more complex, and requires a more thorough analysis. A thermal analysis is outside the scope of this investigation, but experimental investigations have been carried out by other researchers. Some of the published work seems to be in line with the experimental results found in this investigation. For instance, an investigation into the geometrical effect in the microstructure of Ti-6Al-4V produced by L-PBF (TED of 318 J/m) [57] found that parts with a cross-sectional change along the build direction show significant microstructural evolution. In larger cross sections, higher concentrations of vanadium (characteristic of the β phase of this alloy due to martensitic decomposition) could be found, evidencing a localized heat accumulation and lower cooling rates. The idea that a smaller cross-section dissipates heat faster, coupled with the lower heat input present in the core compared to the head of the samples, would serve as further evidence that the local melting mode will be altered, as evidenced in Figure 5.8.

However, not all results seem to agree that smaller cross sections lead to more heat dissipation. A study on the thermal behavior of AlSi10Mg rectangular pieces produced by L-PBF [58] found that the heat flux per unit area in samples with smaller dimensions was greater, both to the surrounding powder and to the build plate. However, the overall heat transferred was greater in larger samples. Another investigation into the size effect in rectangular specimens of NiTi shape memory alloys [59] suggests that decreasing one dimension led to the suppression of the heat flow into the powder parallel to the diminished dimension. Therefore, heat tends to accumulate within the smaller sample, leading to keyhole porosities. This is illustrated in Figure 5.9.

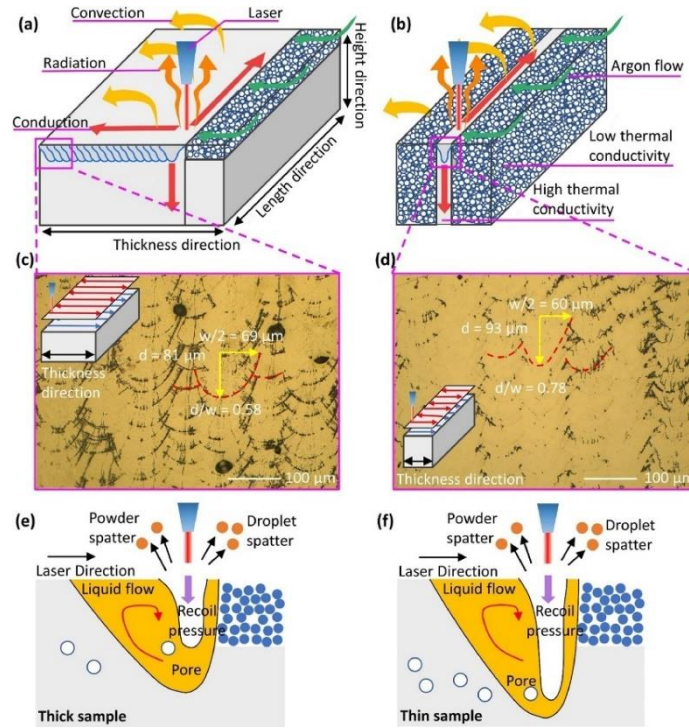


Figure 5.9. Size effect on the heat transfer (a and b), melt pool (c and d), and melting mode (e and f). [59]

From this discussion, it becomes clear that further work needs to be done to characterize the impact of geometry and size on the heat-transfer mechanisms in L-PBF. Nevertheless, the geometry effect on the thermal history is significant, and will affect the melt pool and formation of porosities and surface roughness.

5.3. Impact of Post-Treatment and Process Parameters on Microstructure and Static Properties

The tensile tests results presented in Section 4.3 provide information about the material's strength, yield behavior, and ductility. These results indicate that there are differences between the as-built (AB) and stress relieved (SRHT) conditions, mainly in the ultimate tensile strength (UTS) and the fracture strain. While in AB sample, the UTS hovered around 450 MPa, for SRHT samples it decreases to approximately 410 MPa. The fracture strain is decreased by approximately 1% across all sample groups, going from 4.5% to 3.5% on average. Despite this difference, it is within the margin of error of the measurements. In all conditions, the yield strength ranged from 235 to 255 MPa. These values are comparable to those reported in previous investigations [6, 11, 52]. The different contour parameters do not have a significant impact on the tensile properties. In contrast, the SRHT has an impact on the UTS of the samples. Taking the results from SB4 as a reference, all conditions are superposed in Figure 5.10.

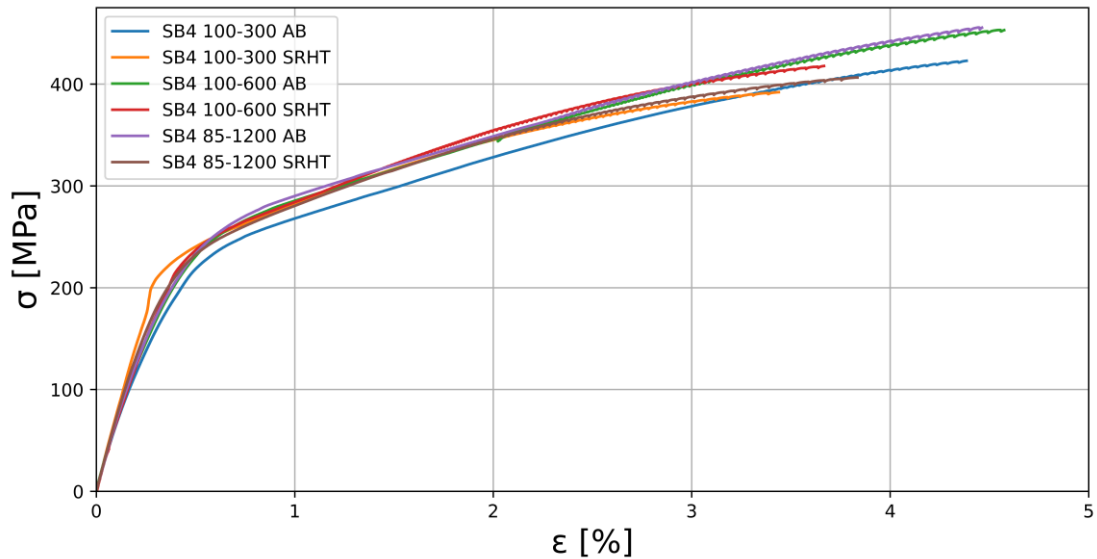


Figure 5.10. Comparative stress-strain curve for AB and SRHT 100-300, 100-600, and 85-1200 samples.

The differences in UTS between the SB and SRHT conditions could be explained from the differences in microstructure. Due to the nature of the L-PBF process, the finished parts have a multiscale and hierarchical microstructural features that will dictate the mechanical behavior. These features include the melt pools and grains at the mesoscopic level, the Al cell and Si network at the microscopic level, and the Si-rich particles and precipitates at the nanoscopic level. This hierarchical microstructure is schematically illustrated in Figure 5.11a, while Figure 5.11b presents an overview of the melt pool hierarchy in an AB 100-600 half sample. The Si network and Al cell can be seen in Figure 4.11.

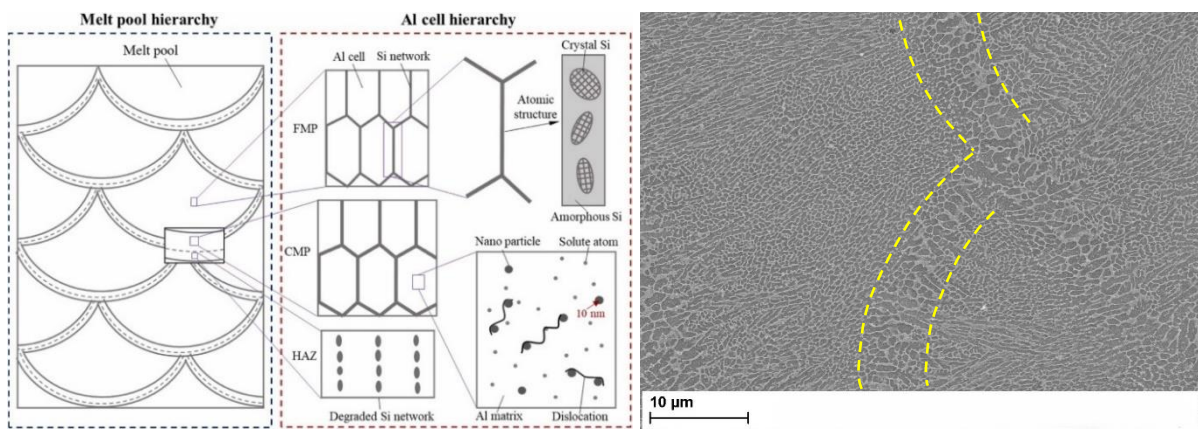


Figure 5.11. a) Schematic representation of the multiscale hierarchical microstructure of AlSi10Mg [60] and b) overview of the melt pool hierarchy in an AB 100-600 half sample.

One possible explanation for the drop in UTS observed in the SRHT samples compared to the AB samples is the globularization of the Si network by the heat treatment. During processing, the high cooling rate associated with L-PBF will create an ultrafine, dendritic Si interconnected network that encloses the Al cell, which imparts the alloy with high mechanical strength in the AB condition. However, an investigation into the effect of post-treatments at different temperatures on the damage mechanisms of AlSi10Mg [52] showed that when samples are heat treated above 300 °C, the Si network will begin to break down and globularize which will increase ductility, but at the expense of the strength as illustrated in Figure 5.12.

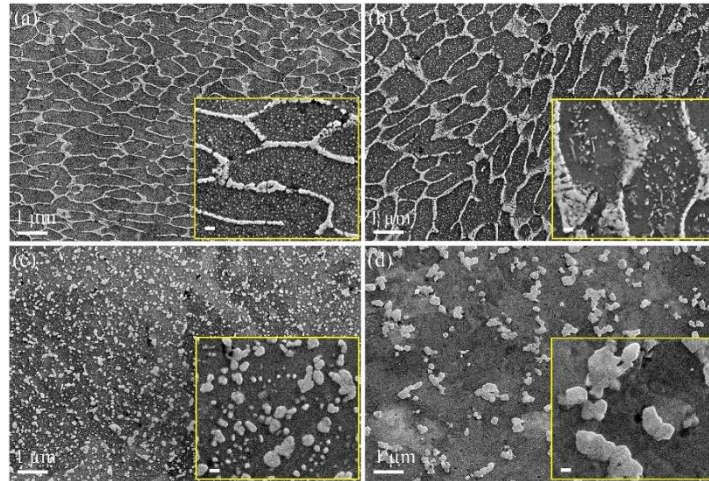


Figure 5.12. Microstructure of AlSi10Mg: a) AB, b) SRHT at 250°C, c) SRHT at 300 °C, d) FSP. [52]

A microstructural comparison between the AB and SRHT condition was presented in Figure 4.11. The fine interconnected Si network in the bulk in the AB (Figure 4.11a) and SRHT (Figure 4.11b) condition showed no significant differences. In contrast, the coarser Si network in the contour in the AB condition (Figure 4.11c) was globularized more after the SRHT (Figure 4.11d). Overall, the heat treatment at 250 °C for 2 hours does not lead to major alterations in the Si network in the bulk, and some globularization and coarsening in the contour. This would explain why a drop in the UTS is not accompanied by a substantial increment in ductility.

Within the sample groups, the effect of the process parameters on the Si network could come from the effect of the energy density and cooling rate. Studies have shown that in AlSi10Mg, the cooling rate can be considered an indicator of the Al cell size, as there seems to be a close correlation between the two at high TED levels. While an investigation into the size change of the Al cell in the conduction melting mode found that decreasing the TED had no impact on the Al cell size [55], other investigations show that a higher cooling rate will promote the growth of a finer cell structure. Furthermore, experimental results show that a higher laser power or lower scanning speeds will lead to a coarser Si-rich network [60]. This behavior is consistent with the Rosenthal 2D equation, discussed in Section 5.1, as the cooling rate is inversely proportional to the TED. This would explain the coarser Si network observed in the contour compared to the bulk in the AB 100-300 sample, as illustrated by Figure 4.11c.

In order to understand the effect of the Al cell size and the Si network on the strengthening of the alloy, it is possible to compare the Si networks that encompass the Al-matrix to grain boundaries. The Al-matrix inside each cell will accommodate plastic deformation by the dislocation of atomic planes, and the density of these dislocations will increase until further movement is restricted by the Si-rich network that acts as a barrier between adjacent cells. Therefore, a coarser Si network coupled with a larger Al cell size could contribute to the slightly decreased UTS in the 100-300 and 100-600 samples in AB conditions. This, however, does not explain the seemingly scattered behavior of the hardening capacity after the heat treatment observed in the SRHT samples. An alternative explanation for this is the connection between the contribution of the Si precipitates within the Al cell to the hardening capacity and their size and shape. Research has shown that this contribution is more significant when these are smaller. However, the size of these particles shows no correlation with the cooling rate or the TED [60], and seems to be more closely connected to the thermal history and precipitation behavior after solidification. Therefore, it is possible that after

the heat treatment, the load-bearing contributions of these precipitates change, thus provoking the scatter observed.

One final aspect to consider are the differences observed in the fracture behavior of the bulk and the contour, and the way they highlight the effect of the keyhole porosities on the mechanical properties. The fracture surfaces for the 100-300, 100-600 and 85-1200 SRHT samples were presented in Figure 4.13. In both the 100-300 and 100-600 samples, keyholes are visible along the fracture plane. Remember that, for tensile samples with a diameter of 4 mm, keyholes are visible in the contour of both the 100-300 and 100-600 samples, possibly due to the geometry effect. Furthermore, under a greater magnification, is possible to distinguish differences between the topology of the bulk and contour regions in the 100-300 samples, as evidenced in Figure 5.13a.

In Figure 5.13b, it is possible to observe that the bulk is marked by an irregular surface characterized by ridges and flat areas. These results are similar to those observed by [55], and indicate that the fracture surface deformed plastically before final fracture, which is consistent with the dimples observed at a higher magnification (not shown). The flat areas in between the ridges present some concavity and suggest that the fracture occurred at the border of the melt pool, in the Heat Affected Zone (HAZ), illustrated in Figure 5.11a. Unlike the Fine and Coarse Melt Pools (FMP and CMP respectively), the HAZ does not present a fine interconnected Si network that encloses the Al cell (see Figure 5.11b) exhibiting instead a degraded globularized network that is formed by an intrinsic heat treatment since the temperature in this region does not exceed neither the liquidus nor the solidus temperatures. Investigations suggest that the HAZ at the melt pool borders is the preferred fracture location in vertically-built specimens [60], like the ones used in this work.

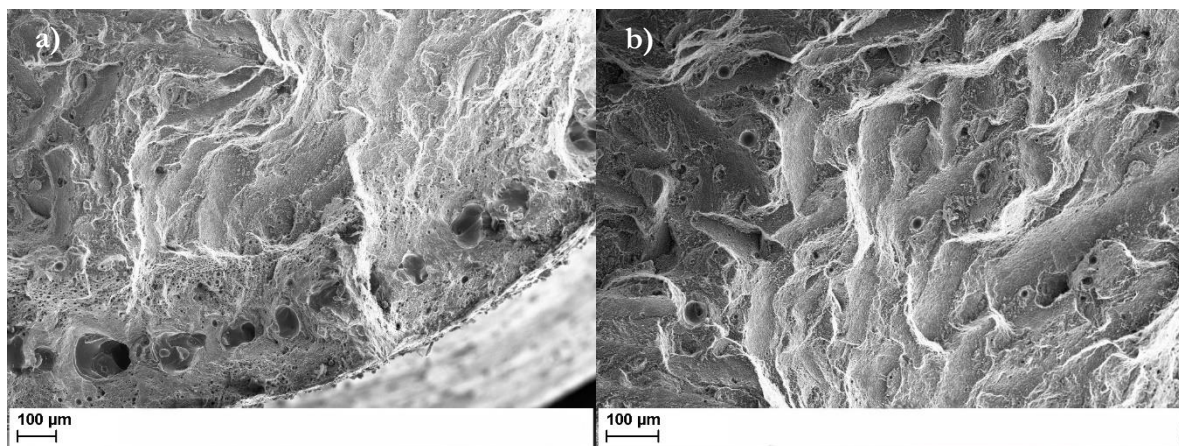


Figure 5.13. a) Contrast between fracture surface at the bulk and contour and b) fracture surface at the bulk of a SRHT 100-300 sample.

Unlike the bulk, the contour does not present the ridges or characteristic flat surfaces that would indicate that the fracture initiated at the border of the melt pool. This is consistent with the coarser Si network in the contour observed in Figure 4.11d. Since the Si network is coarser and globularized, the fracture location will not necessarily be confined to the melt pool boundary. The presence of keyholes all along the fracture plane is evident. This suggests that the reduced load bearing capacity of the contour due to the porosities has a higher impact on the fracture behavior than the microstructure of the melt pools. Overall, the heat treatment reduces the UTS, but not the yield stress, suggesting that the microstructure in the AB and SRHT conditions is similar.

5.4. Impact of Process Parameters and Post-Treatments on Fatigue Properties

The fatigue performance of net-shaped, vertically-built AlSi10Mg fatigue specimens with different contour parameters was presented in Section 4.5. The Wöhler curve (Figure 4.15) reveals that the increased contour TED in the 100-300 and 100-600 samples results in a better fatigue performance than the reference 85-1200 samples. The 100-300 and 100-600 samples present a lower surface roughness than the 85-1200 samples (Figure 4.8), while only the 100-300 samples present keyhole porosities along the contour. This indicates that roughness rather than microstructure or porosity is the controlling factor for fatigue performance in the studied stress range. Moreover, the results suggest that keyhole porosities do not represent a detrimental factor for fatigue performance, since SRHT 100-300 samples presented the best overall fatigue life.

In order to compare the performance of the SRHT sample groups to other experimental results with a similar building and testing conditions reported in the literature, the results from the Wöhler curves are fitted using Basquin's Law:

$$\sigma_a = C N_f^b$$

where σ_a represents the stress amplitude, N_f represents the cycles to failure, and the constants C and b represent the fatigue strength's coefficient and exponent, respectively. The fatigue strength exponent represents the Wöhler curve's slope on a log-log scale and is related to the material's sensitivity to cyclical loading: a higher value of m implies a more severe decrease in fatigue life for a given increase in stress amplitude. On the other hand, the fatigue strength coefficient represents the intersection with the stress amplitude axis and is related to the fatigue strength of the material: the higher the magnitude of C , the higher the stress amplitude needed to cause failure at a given number of cycles. The calculated values for m and $\text{Log}(C)$, along with the information on the SRHT, the contour TED, and the average arithmetic roughness (R_a) are presented in Table 5.2 .

Sample Group	Post-treatment	Contour TED [J/m]	R_a [μm]	m	$\text{Log}_{10}(C)$
100-300	SRHT (2 hours at 250°C and furnace cooling)	912	4.29	-0.23	3.37
100-600		456	5.23	-0.29	3.64
85-1200		193.8	8.82	-0.23	3.29

Table 5.2. Contour TED, average arithmetic roughness (R_a), and Basquin's Law parameter for SRHT 100-300, 100-600 and 85-1200 samples.

The results from the Basquin's Law fitting mirror the behavior observed in Figure 4.14 and Figure 4.15. Both the 100-300 and 85-1200 samples present an identical slope, indicating that the material sensitivity in both cases is similar, while the 100-600 samples present a higher slope, which could indicate a higher sensitivity compared to the other sample groups. In both the 100-300 and 100-600 samples, the higher $\text{Log}(C)$ values demonstrates the overall improvement in fatigue performance and the shifting of the graph to a higher stress range. Comparatively, the m values of all sample groups agree with the results for as-built samples present in the literature [61, 62, 63], and reaffirms the microstructural similarities among these samples. An important difference is observed in the range of $\text{Log}(C)$ values reported in the literature for AB samples, as these are lower than those obtained here for all SRHT sample conditions, being in the vicinity of 2.9.

The fatigue performance of the 100-300, 100-600, and 85-1200 samples indicates that even a non-roughness-optimized contour process window leads to a better fatigue performance than AB samples without contour. To evaluate the fatigue performance achieved by optimizing contour parameters, it is necessary to compare it to the effect of different post-treatments. In order to improve the fatigue performance, previous investigations [6, 46, 47, 61, 62, 63, 64] have focused on a post-treatment or a combination of post-treatments which have at least one of the following objectives:

- Alter the microstructure in order to increase ductility, refine grain morphology, and provide a better resistance to crack propagation, which involves a heat treatment
- Improve the surface quality by altering the topology and reducing the propensity for crack nucleation, which implies removal of the outermost layers of the part through mechanical or manual means.
- Induce residual compressive stresses on the surface in order to reduce the probability of crack nucleation, which implies creating microscopic plastic deformations at the surface and sub-surface level of the part through an external application of force.

Investigations on the fatigue properties of AlSi10Mg have shown that, in general, the objective that is achieved will make the nature of the improvements in fatigue performance vary. In particular, improvements in surface quality tend to shift the curve to a higher stress range (reflected by a higher Log (C)), while alterations in the microstructure and the introduction of compressive residual stresses will decrease the material sensitivity to stress (reflected in a reduction in the magnitude of m).

For example, an investigation focusing on the effect of heat treatment and surface quality [61], found that machined specimens had a Log (C) coefficient of 3.1 compared to the AB 2.9 reference value. However, this improvement was limited. In contrast, the use of a T6 treatment decreased the magnitude of the m exponent to -0.15 compared to an initial -0.22. This is illustrated in Figure 5.14. In another investigation [47], failure in machined specimens for a given fatigue lifetime occurred at a stress 20 MPa higher compared to the AB condition, while using a T5 treatment allowed to decrease the sensitivity of the m exponent, as seen in Figure 2.22.

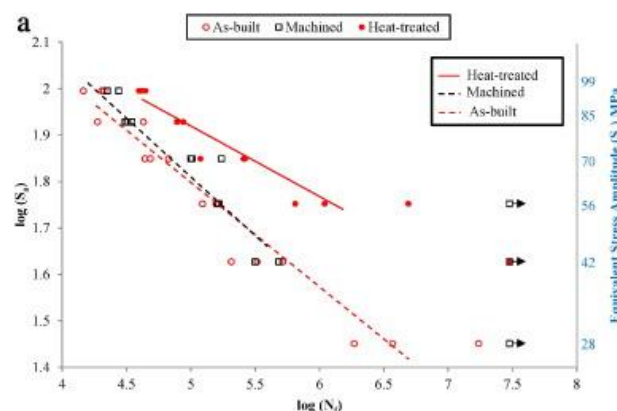


Figure 5.14. Fatigue performance of as-built, heat-treated and machined samples. [61]

One probable cause for the improvement in fatigue performance achieved by optimized contouring techniques can be the reduction in surface roughness. In order to illustrate the degree of improvement compared to traditional surface quality-improving techniques, the SRHT results are compared to the polished (P) ($R_a < 1 \mu\text{m}$) samples from the investigation carried out by Santos et al. [6] by plotting the normalized Wöhler curves in Figure 5.15.

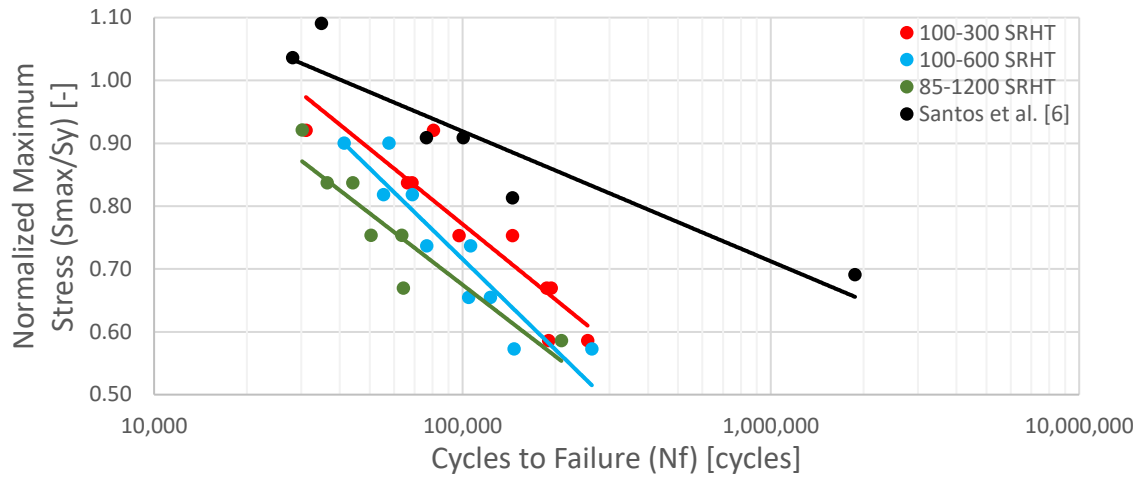


Figure 5.15. Normalized Wöhler curves for SRHT 100-300, 100-600, 85-1200, and P samples from Santos et al. [6].

Aside from the outlier value at the lowest stress level (which the authors indicate was carried out using a different fatigue machine), the trend for the P samples is similar to that of the roughness-optimized 100-300 and 100-600 samples. Furthermore, these results seem to indicate that the presence of keyholes in the contours of the sample do not have a detrimental effect on the fatigue performance of the SRHT 100-300 samples.

The roughness measurements show that optimized parameters results in a lower roughness and better fatigue performance. To evaluate possible improvements in fatigue life from further post-processing, the 100-300 and 100-600 specimens were polished. The purpose of this step is to assess on one hand, how much fatigue life could be improved by a further improvement in surface quality, and in the other, the independent effect of keyholes in the 100-300 samples compared to the 100-600 (without keyholes). The results presented in Figure 4.16 and Figure 4.17 evidence that overall, polishing has a positive impact in the fatigue life, by shifting the graph to a higher stress range for a given fatigue lifetime and decreasing the sensibility to cyclic loading. These results are compared to those obtained by Santos et al. [6] and presented in Figure 5.16.

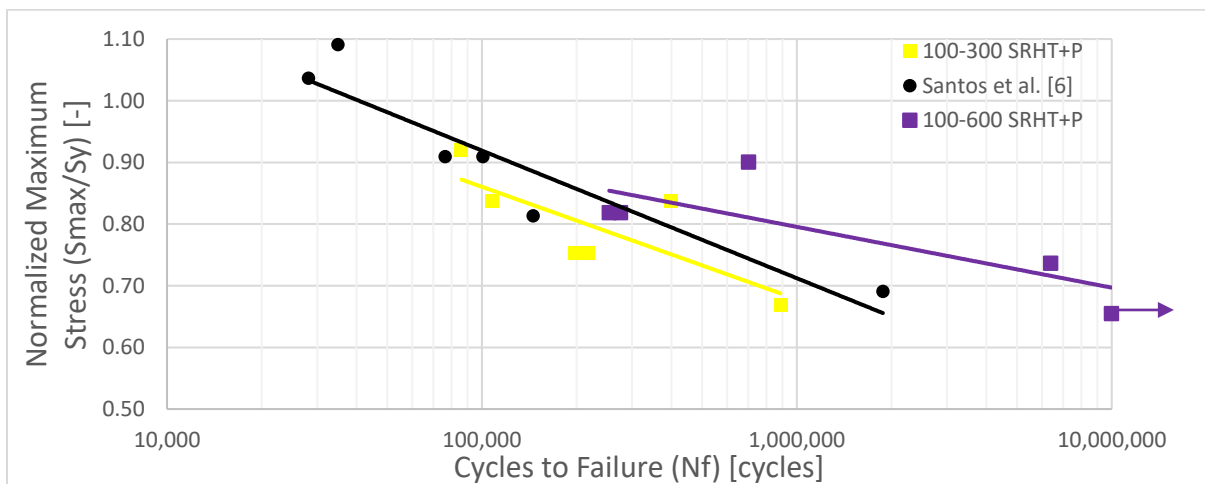


Figure 5.16. Normalized Wöhler curves for SRHT+P 100-300 and 100-600 samples and P samples from Santos et al. [6].

By polishing to achieve an average surface roughness $R_a < 1 \mu\text{m}$, it is possible to improve the fatigue life even further compared to the SRHT condition and reach a comparable and in some cases better performance compared to Santos et al. [6]. Notice that the performance of the SRHT+P 100-600 samples appears to be better than that of the SHRT+P 100-300 samples, unlike the SRHT case. The reason behind this difference comes from the effect of the keyhole porosities on the crack nucleation at the polished surface of 100-300 samples, as shown in Figure 5.17.

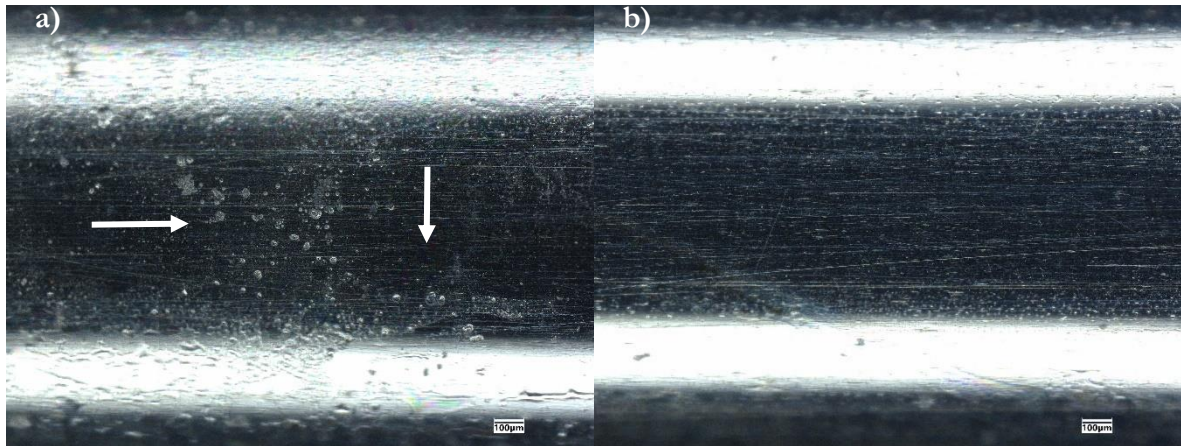


Figure 5.17. Comparison of the polished surface of the a) 100-300 and b) 100-600 samples.

By polishing the surface following the steps described in Section 3.4.2, approximately 50-100 μm were removed from the surface. This removal of the outer layers improves the overall state of the surface, as any attached unmelted powder particles are removed, and the grooves on the surface are likewise extinguished. However, as the material is removed, these keyhole porosities in the contour are exposed, creating potential hotspots for crack nucleation. This does not happen in the case of the 100-600 samples because, by operating at the onset of the keyholing mode, there are almost no keyholes at the contour as evidenced in Figure 5.17b. Therefore, polishing is more efficient at reducing roughness in SRHT 100-600 samples, making them more immune to crack nucleation. This is consistent with the fatigue results, where only the SRHT+P 100-600 sample group had a runoff in the stress range in which the samples were tested.

The effect of machining on the opening of sub-surface defects has been noted in other investigations. In [65], Nasab et al. found that by applying vibro-finishing to fatigue specimens, it was possible to increase their fatigue lifetime. However, sub-surface defects produced during the L-PBF process had been opened. The authors found that the defects that were exposed to the surface after the post-treatment were mainly lack of fusion defects. In this work, the exposed defects are mainly keyholes. Despite the fact that the exposure of any sub-surface defect is detrimental for fatigue performance because it constitutes a preferential site for crack nucleation, the irregular shape of a lack of fusion porosity will induce a higher stress concentration compared to a keyhole porosity with a relatively high sphericity.

Despite improving fatigue life by improving surface quality, Aboulkhair et al. [61] argued that using this strategy creates a non-uniform improvement across the different stress levels. This is because fatigue life is dominated by the crack growth behavior at high stress level and by crack nucleation at low stress levels. Therefore, any improvements with respect to an AB condition will be mainly observed at lower stress levels. This is consistent with the comparison of SRHT and SRHT+P 100-300 and 100-600 samples (Figure 4.16 and Figure 4.17), where the improvement of SRHT+P samples was more significant at lower stress levels.

Based on this, Aboulkhair et al. [61] proposed that to achieve a uniform improvement across all state levels (same m , higher $\text{Log}(C)$), it is necessary to induce microstructural change via a heat treatment. Their results indicate that the combination of a T6 heat treatment coupled with machining operations produced the best overall results (Figure 5.14). The arguments presented are founded on the crack growth behavior and the way the Basquin parameters evolve after machining/polishing. The fracture surface analysis at 220 and 140 MPa presented in Figure 4.21 coincides with the results obtained by Aboulkhair et al. [61]. As the stable crack growth region decreases in size as the maximum stress decreases (Region 2) while the inverse is true for the initiation and slow crack growth region (Region 1).

Another investigation [47] also reported improved results by combining a T5 treatment with machining. Furthermore, from the results presented in Figure 4.15, Figure 4.17, and Figure 5.15, it is possible to see that at higher stress levels, the fatigue lifetimes achieved by the P samples are close to those achieved by the SRHT samples, and that the most significant improvements occur at lower stress levels. However, some results have shown that the addition of a heat treatment, particularly a T6 heat treatment, might be detrimental for fatigue life. Both [6] (Figure 2.19b) and [66] (Figure 5.18) report that the globularization of the Si network at the boundaries of the Al cell causes the fatigue performance to drop.

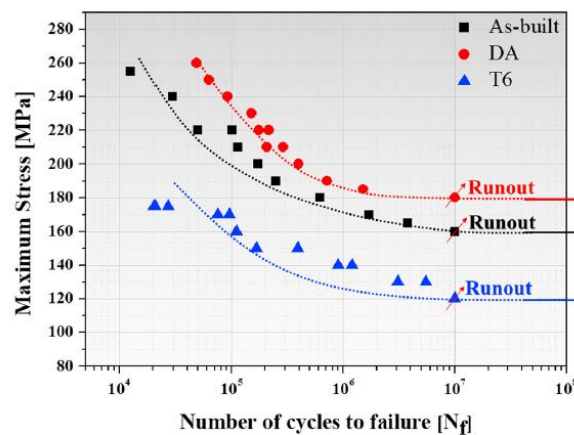


Figure 5.18. Effect of a T6 heat treatment on fatigue performance [66].

The use of post-processing that causes localized microstructural change has been explored as an alternative to heat treatment. Santos et al. [6] reported significant improvements in fatigue lifetime by applying friction stir processing at various advance speeds (Figure 2.19b). Another option is the use of sandblasting (used in this investigation) and shot-peening, which induce residual compressive stresses by localized plastic deformation of the surface and may reduce roughness. Figure 5.19 presents the evolution of the surface roughness in SRHT and SRHT+SB samples compared to the improvements reported by Avanzini et al. [62]. When contour-optimized parameters are used, the improvement in surface quality is very low, and there may even be a small increase in the surface roughness after SB. This is because the roughness level achieved in the present work by optimizing the contour parameters is already 1.5-2 times lower than that of the AB samples reported in the literature.

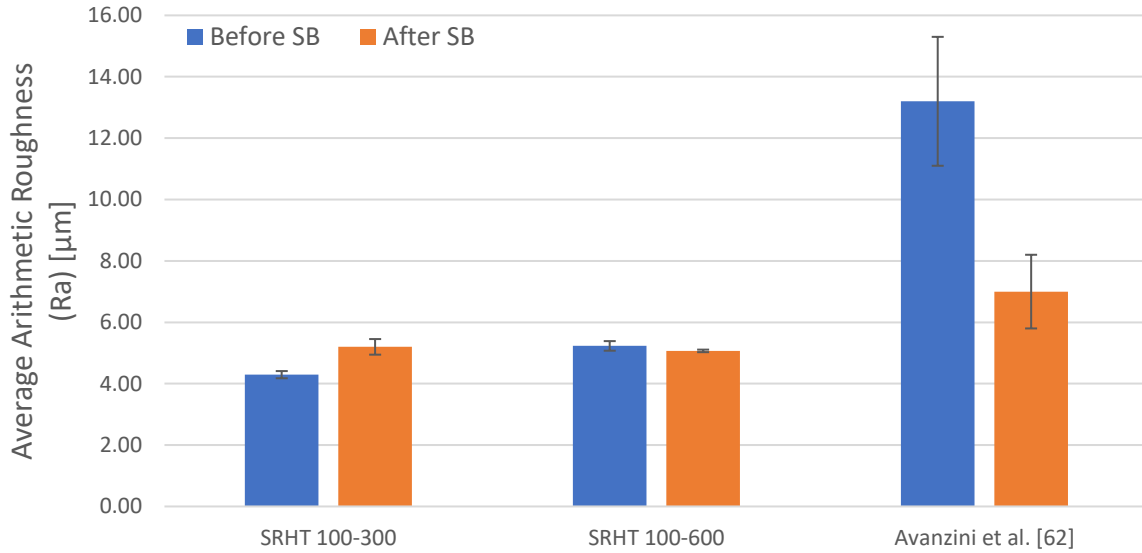


Figure 5.19. Evolution in the Ra before and after sandblasting in SRHT 100-300 and 100-600 samples.

Despite no significant improvement in the surface roughness, the introduction of compressive residual stresses into the surface of the samples is evidenced by the fracture surface of a SRHT+SB 100-600 specimen, as illustrated in Figure 5.20. The red line in both pictures separates the zone visually affected by the sandblasting in the contour region of a SRHT+SB 100-600 sample. In this zone, the Al cells are compressed together, leading to an agglomeration of the Si network boundaries that enclose each cell (Figure 5.20b). It is also possible to see that plastic deformation caused by sandblasting spans approximately 0.1 mm from the surface.

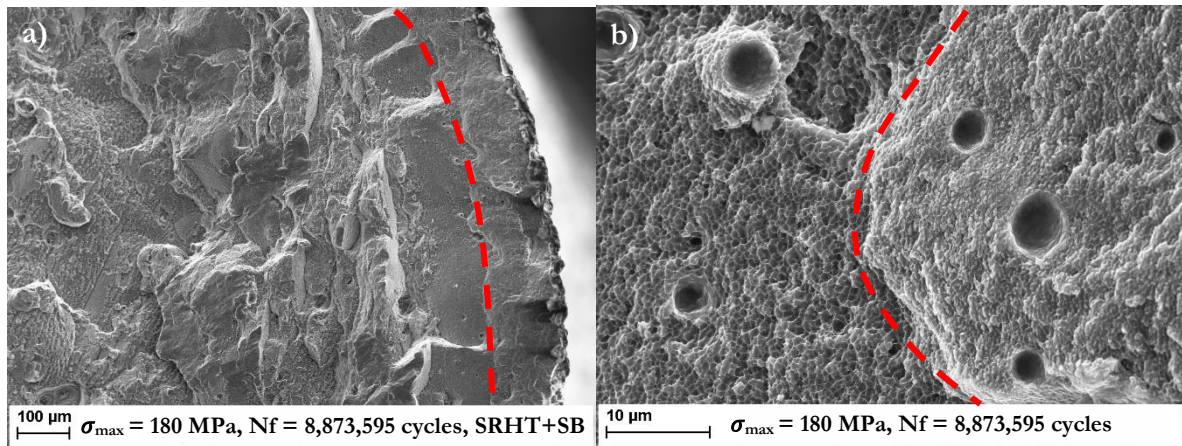


Figure 5.20. Influence of SB on the contour microstructure of a SRHT+SB 100-600 sample at a) 100x and b) 2000x magnification.

The performance of SRHT+SB 100-300 and 100-600 samples compared to the SRHT+P group is presented in Figure 5.21. The improvement in fatigue lifetime achieved by SB compared to the P condition is evident in both the 100-300 and 100-600 samples and can be attributed entirely to the introduction of residual compressive stress. A large scattering at high stress levels can be observed. Due to the high cooling rates associated with L-PBF, samples often have significant tensile residual stresses at their surface. This deteriorates the fatigue performance, as it contributes to crack nucleation and propagation. The heat treatment performed on the SRHT samples likely contributed to relax the stresses and may have contributed to improvement in fatigue life. Nevertheless, the improvement achieved by sandblasting is more significant.

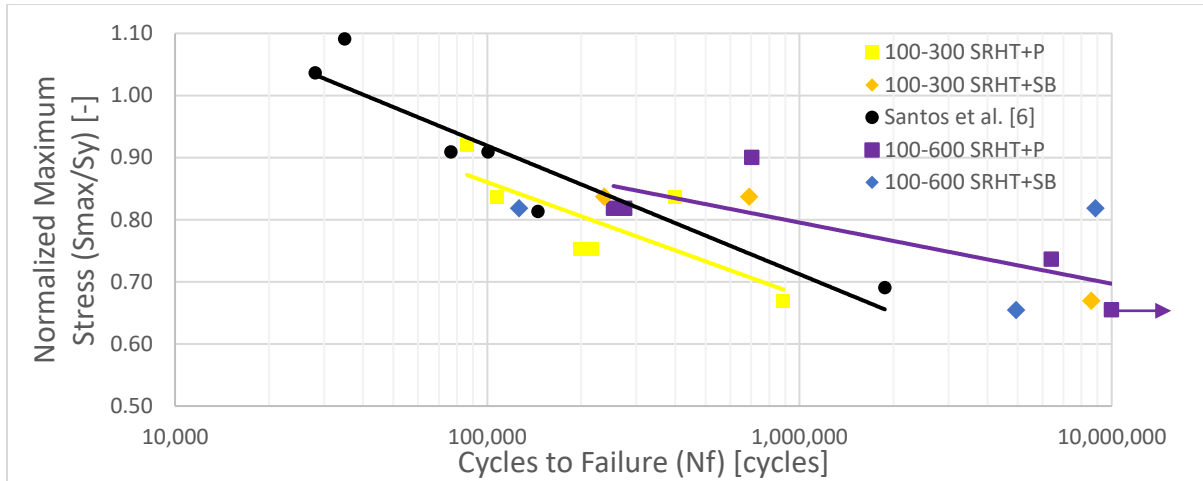


Figure 5.21. Normalized Wöhler curves for SRHT+P and SRHT+SB 100-300 and 100-600 samples and P samples from Santos et al. [6].

Bagherifad et al. [64] reported that by sandblasting AlSi10Mg samples, it was possible to go from a residual stress of 70 MPa to -90 MPa at the surface of the sample and up to a peak stress of -145 MPa at a depth of 0.12 mm. These results have been reproduced by [62], who reported that the fatigue strength exponent (m) was decreased to approximately -0.13, marking an improvement over the AB condition. However, one of the main drawbacks associated with sandblasting identified in this investigation is the lack of repeatability. Figure 4.23 illustrates the difference in the crack nucleation sites of 100-300 and 100-600 SRHT+SB samples. Two samples with the same contour parameters subjected to the same sandblasting process tested at the same stress level can have vastly different lifetimes. While in the shortest-lived samples the initiating defect was at the surface, in the longest-lived samples the initiating defect shifted to an internal pore in the bulk region. This highlights the positive impact SB can have in hardening the surface, but also that local variations in the surface might lead to hotspots for crack nucleation.

6. Conclusions

The presence of porosities and low surface quality of samples manufactured by L-PBF have a detrimental effect on the mechanical performance, especially under cyclical stress. In order to guarantee a reliable performance in engineering applications without the use of time-consuming post-treatments, efforts have been directed towards the optimization of the L-PBF. In this context, this investigation aimed to study the influence of roughness-optimized contour parameters on the fatigue behavior of net-shaped, vertically-built AlSi10Mg samples. To achieve this, samples were produced with a contour-first strategy using a laser power of 232.5 W and a scanning speed of 1200 m/s for the bulk. A parametric study was carried out to select three different contour process windows for 3 mm core diameter fatigue samples. The following were selected:

- 100-300 Samples: Laser power of 273.6 W and a scanning speed of 300 mm/s, corresponding to keyhole melting mode.
- 100-600 Samples: Laser power of 273.6 W and a scanning speed of 600 mm/s, corresponding to the onset of the keyhole melting mode.
- 85-1200 Samples: Laser power of 232.5 W and a scanning speed of 1200 mm/s, corresponding to the reference case (same process parameters in bulk and contour).

First, a characterization of the effect of geometry and process parameters on the surface roughness and keyhole porosities was carried out. Next, the microstructure and tensile properties of 100-300, 100-600, and 85-1200 samples were studied before (as-built, AB) and after (heat treated, SRHT) a heat treatment at 250°C for 2 hours. Finally, fatigue properties of SRHT samples from all groups were compared to that of polished (SRHT+P) and sand blasted (SRHT+SB) samples to evaluate the impact of an improved surface quality and contour keyhole porosities.

A comparative parametric study on the influence scanning speed for a fixed laser power reveals that, for the same track energy density (TED), the geometry of the sample will impact the melting mode. The arithmetic surface roughness (R_a) and presence of keyholes at the contour will vary along the building direction of a cylindrical fatigue specimen and will depend on the diameter of the cross section. Moreover, using samples with a larger diameter at the core caused a decrease in surface roughness and the need for a lower TED to achieve keyhole conditions. This was evidenced by using samples with 3 instead of 2 mm at their core, in which a reduction of 20 – 40 % in surface roughness was observed. Scanning speeds of 300 mm/s (TED: 912 J/m) and 600 mm/s (TED: 456 J/m) were enough to be in full and at the onset of keyhole mode, respectively, while simultaneously achieving surface roughness levels below 5 μm . The total heat input and possible lower heat dissipation rate associated to larger cross-sectional areas have been identified as the main contributors to the geometry effect.

A contour process window with a TED that causes conditions close to and in full keyhole melting mode results in an improved surface quality and the creation of keyhole porosities. Regarding roughness, a higher TED causes a lower cooling rate, a high temperature latency that promotes good wettability (reducing viscosity and surface tension), and flat edges of melt pools. These factors inhibit the adherence of unmelted powder particles, thus reducing the peak roughness. This was evidenced by a reduction of the surface roughness from 8.8 μm in the 85-1200 samples, to 4.2 μm and 5.1 μm in the 100-300 and 100-600 samples. Regarding keyhole porosities, by operating at the onset and in full keyhole mode, the dimensions of the melt pool are increased. This is evidenced by an increase in the contour track and depth-to-width ratio for an increasing TED in the 100-300 and 100-600 samples. The increase in the depth-to-width is related to a melt pool

instability that causes vaporized material bubbles to be trapped, creating keyholes with a high sphericity and smaller trapped-gas porosities at the outer and inner edges of the contour.

The SRHT at 250°C for 2 hours did not significantly alter the microstructure of the AB specimens. An interconnected Si network was observed in the bulk before and after the heat treatment, with no evidence of widespread globularization. In the contour, a higher TED created a coarser Si network, which was more deeply affected by the heat treatment. Another effect of the heat treatment is the decrease in the ultimate tensile strength (UTS) observed in the SRHT samples compared to the AB samples. In contrast, the process parameters do not have a major impact on the static performance of the SRHT or AB specimens. While the fracture surface revealed that the fracture plane is marked by the presence of keyholes in the 100-300 and 100-600 samples, no major differences in UTS or ductility were observed. This is explained because unlike lack of fusion porosities, keyholes have a high sphericity, and thus do not concentrate effort as much. After the SRHT, the scattering in the UTS can be attributed to the contribution of Si precipitates within the Al cell.

The effect of the contour parameters in the fatigue performance is positive, particularly when these are printed to shape. Both the 100-300 and 100-600 sample groups performed better than the reference 100-30

0 samples, and the performance of 100-300 is comparable to that of polished samples at high stress levels. This improvement is mainly due to the reduction of surface roughness achieved in the 100-300 and 100-600 samples obtained by operating at the onset and in full keyhole mode. Despite the presence of keyhole porosities, these do not affect fatigue performance in the studied stress range. Further improvements can be accomplished by polishing and sandblasting. When polishing was applied, SRHT+P 100-600 samples performed better than SRHT+P 100-300 samples, likely due to the fact that polishing removes the outer layers of the contour and exposes the keyhole porosities. Sandblasting produced the samples with the best performance. SHRT+SB 100-300 and 100-600 samples surpassed the SHRT+P samples in this work and some present in the literature. However, one negative effect is the variability inherent to the process, as the same post-treatment might lead to different crack initiation mechanisms at the same stress level.

In summary, by identifying the optimum process window for the geometry under consideration, using a contour strategy at a TED level at the onset or in full keyhole mode will improve the fatigue behavior of AlSi10Mg. In net-shaped samples, operating in keyhole conditions provides the best results because of the low surface roughness, and the keyhole porosities in the contour play no role in the fatigue mechanisms. In polished samples, the exposure of keyholes might prove detrimental to the fatigue behavior, as they may constitute sites for crack nucleation, and thus operating at the onset of keyhole provides better results. Sandblasting produced the samples with the longest fatigue lifetimes, despite no visible improvements in surface roughness, due to the introduction of significant compressive residual stress. It constitutes a less time-consuming alternative to polishing, making it an attractive alternative if ways to standardize the process and improve repeatability are investigated. Further improvements in fatigue performance might be achieved by exploring combinations of optimized contour parameters with heat treatments that reduce crack propagation.

7. References

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10. Annexes

10.1. Preliminary Keyhole and Surface Roughness Results of Batch 1

The process parameters used to manufacture Batch 1 were based on the results of Poncelet et al. [10] and Mache [11] in order to evaluate contour parameters at the onset of keyhole (“cold” contour) and in keyhole (“hot” contour) conditions, using scanning speeds of 700 and 400 mm/s, respectively. A no contour reference case using a Type A fatigue specimen geometry has been used for comparison (illustrated in Figure 3.1). Table 10.1 summarizes the TED level in the bulk and contour for each of these cases.

Sample Designation	Bulk			Contour		
	Laser Power [W]	Scanning Speed [mm/s]	TED [J/m]	Laser Power [W]	Scanning Speed [mm/s]	TED [J/m]
C-100-400 (Hot Contour)	232.5	1200	193.8	273.6	400	684
C-100-700 (Cold contour)					700	390.9
No Contour	-	-	-	-	-	-

Table 10.1. TED of the “hot” contour, “cold” contour, and reference (no contour) cases in Batch 1.

Initially, the C-100-400 specimens were expected to exhibit keyhole porosities in the contour region, while the C-100-700 specimens were expected not to show any. However, Figure 10.1 depicts the cross-sections at the core and no significant keyhole porosities with a size comparable to that of the melt pools are visible along the contour in either case. Conversely, the cross-sections at the heads of the specimens (with a nominal diameter of 5 mm) exhibit keyhole porosities in the case of the C-100-400 specimens, and track instabilities in the case of the C-100-700 specimens, as observed in Figure 10.1c and d. Microstructural differences may be expected within samples produced in a layer-wise process due to the differences in thermal history between the different layers. However, the fact that keyhole porosities are not present at the core suggests that the heat transfer mechanisms change along the building direction and are linked to the geometry of the cross section.

The lack of porosities in the core region of the C-100-400 samples was not the only unexpected result. Based on the two previous investigations on the subject [10][11], a roughness level (R_a) below 6 μm was expected. However, as in the case of the keyhole porosities, significant differences were observed at the head and core sections of the samples, as illustrated in Figure 10.2. Using optimized parameters produces roughness levels in the same order of magnitude as those in the literature only in the head sections of the C-100-400 and C-100-700 specimens. In contrast, the core sections have roughness levels which are twice as high and are comparable to the use of non-optimized process parameters. The roughness results are consistent with the contour porosity distribution and reinforce the notion that geometry plays a role in the heat transfer mechanisms during the L-PBF process. It is important to highlight that, visually, the roughness in the core section of the sample seemed to vary along the sampling length, which means that the resulting value considers a reduced roughness as the cross section becomes wider near the start and end points of the sampling length, and a higher roughness at the geometrical center of the sample. These differences, however, were not able to be captured with the roughness measurements, as they are very localized.

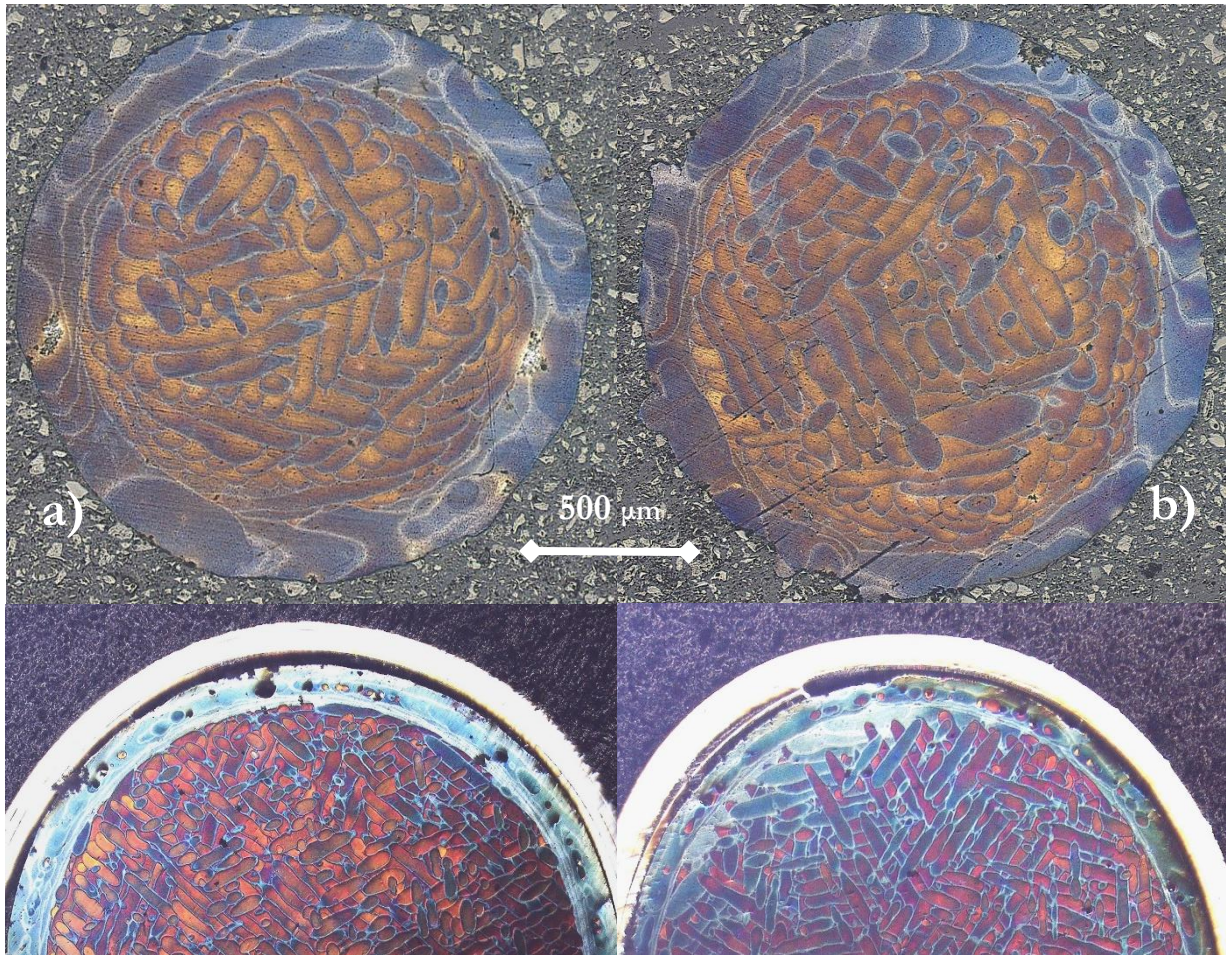


Figure 10.1 Cross sectional view of C-100-400 at the a) core and c) head and of C-100-700 at the b) core and the d) head.

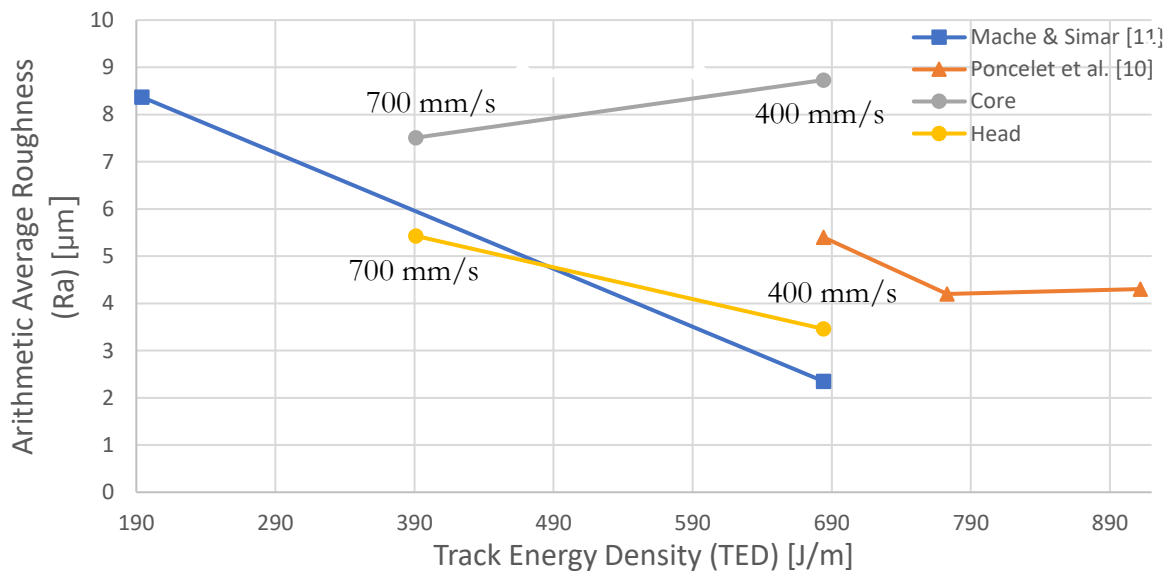


Figure 10.2. Comparison of the surface roughness (Ra) at the head and core sections of the fatigue samples.

10.2. Sample manufacturing summary per batch

Batch 1 (X200_240118)					
Specimen Type	Geometry	Scanning Strategy	Contour		No. of Samples
			Laser Power [%]	Scanning Speed [mm/s]	
Fatigue	Net Shaped	Contour First	100	400	10
				700	10
	Full Cylinder	No Contour	-	-	8
		No Contour	-	-	6
Tensile	Net Shaped	Contour First	100	400	2
				700	2
				No Contour	-
Batch 2 (X200_240220)					
Specimen Type	Geometry	Scanning Strategy	Contour		No. of Samples
			Laser Power [%]	Scanning Speed [mm/s]	
Fatigue	Half Sample (Type A)	Contour First	100	200	3
				300	3
				400	3
				500	3
				600	3
				700	3
				1200	3
	Half Sample (Type B)	Contour First	100	200	3
				300	3
				400	3
				500	3
				600	3
				700	3
				1200	3
		No Contour	-	-	3
		No Contour	-	-	3
Batch 3 (X200_240226)					
Specimen Type	Geometry	Scanning Strategy	Contour		No. of Samples
			Laser Power [%]	Scanning Speed [mm/s]	
Fatigue	Net Shaped (Type B)	Contour First	100	300	12
				400	4
				600	12
				1200	8
Tensile	Net Shaped	Contour First	100	300	2
				600	2
				1200	2

Batch 4 (X200_240228)					
Specimen Type	Geometry	Scanning Strategy	Contour		No. of Samples
			Laser Power [%]	Scanning Speed [mm/s]	
Fatigue	Net Shaped (Type B)	Contour First	100	300	12
				200	4
				600	12
			85	1200	8
Tensile	Net Shaped	Contour First	100	300	2
				600	2
			85	1200	2

10.3. Fatigue testing campaign description

Contour Parameters		Max. Stress Level [MPa]	Post-processing (# of samples)		
Laser Power [W]	Scanning Speed [m/s]		As Built	Polished	Sand-blasted
100	300	220	2	1	
		200	2	2	2
		180	2	2	-
		160	2	1	1
		140	2	-	
	600	220	2	1	
		200	2	2	2
		180	2	2	-
		160	2	1	1
		140	2	-	
85	1200	220	1	-	
		200	2	-	
		180	2	-	
		160	1	-	
		140	1	-	

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