Fatigue behavior of Ti-6Al-4V alloys manufactured by Selective Laser Melting

Influence of microstructure, surface roughness and surface morphology

Thesis

Utmattningsegenskaper för Ti-6Al-4V legeringar tillverkade av Selective Laser Melting

Inflytandet av mikrostruktur, ytfinhet och ytorientering

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Abstract

The intention of this thesis is to investigate the fatigue behavior of the Ti-6Al-4V alloy manufactured by the relatively new additive manufacturing process called Selective Laser Melting (SLM). SLM has been shown the interest from important industries such as the aerospace and biomedical industry for its potential of manufacturing lightweight and complex parts to net shape without the need of conventional methods like machining.

Due to the short history of SLM there is little knowledge about how fatigue properties are influenced by the manufacturing process. This project aims to add to that knowledge by studying how the microstructure, surface roughness and surface morphology influences the fatigue life of SLM made Ti-6Al-4V alloys. Fatigue tests are done by conducting 3-point bending tests on SLM and conventional samples.

It was found that the SLM samples only managed 9% of the fatigue life obtained for conventional samples. This was assumed to be caused by the poor surface roughness of SLM samples, acting as stress concentrations resulting in multiple crack nucleation leading to final fracture. It was also found that a surface morphology of 90°, relative to the length of the sample, had worse fatigue life compared to 60° and 30° due to the pattern of the sample being perpendicular to the applied load.
Sammanfattning

Syftet med denna studie är att undersöka utmattningsegenskaper för Ti-6Al-4V legeringen tillverkad av en relativt ny additiv tillverkningsmetod kallad Selective Laser Melting (SLM). Stora industrier som flygplansindustrin och biomedicin har visat intresse för SLM för sin potential att producera lätt och komplexa produkter till slutgiltig form utan behovet av konventionella metoder som fräsning.

På grund av sin korta historia finns liten kunskap över hur tillverkningsmetoden SLM påverkar utmattningsegenskaper hos produkterna. Förhoppningen är att denna studie ska addera kunskap om hur mikrostruktur, ytfinhet och ytorientering påverkar utmattningslivet för Ti-6Al-4V legeringar tillverkade av SLM. Utmattningslivet prövades för konventionellt och SLM framställda prover genom 3-punkts böjning.

Resultaten visar att SLM prover endast klarar 9% av utmattningslivet som uppmättes för de konventionellt framställda. Detta antas bero på den grova ytfinhet som erhölls för SLM proverna som ledde till höga spänningskoncentrationer, sprickinitieringar och slutligen brott.

Det framkom också att en ytorientering av 90°, relativt provets längd, hade lägre utmattningsliv än 60° och 30° eftersom provets ytorientering var vinkelrätt placerad i förhållande till den pålagda lasten.
Contents

1. Introduction ......................................................................................................................... 1
   1.1 Applications of lightweight material ................................................................................. 1
      1.1.1 Needs ...................................................................................................................... 1
      1.1.2 Conventional Titanium ............................................................................................. 1
      1.1.3 Conventional Titanium alloys .................................................................................... 2
      1.1.4 Influence of microstructure on mechanical properties of Ti-6Al-4V alloys .......... 4
   1.2 Additive manufacturing ..................................................................................................... 5
      1.2.1 Classification of additive manufacturing methods ....................................................... 5
      1.3 Selective Laser Melting .................................................................................................... 6
         1.3.1 Advantages of Selective Laser Melting .................................................................. 7
         1.3.2 Disadvantages of Selective Laser Melting .............................................................. 8
   1.4 Influence of microstructure on mechanical properties of Selective Laser Melting Ti-6Al-4V alloys ................................................................................................................. 9
   1.5 Fatigue in general ............................................................................................................. 9
      1.5.1 3-Point bending .......................................................................................................... 11
      1.5.2 Fatigue of Selective Laser Melting Ti-6Al-4V alloys ................................................. 12
   1.6 Aims ............................................................................................................................... 13

2. Method .................................................................................................................................. 14
   2.1 Preparation of fatigue samples made by Selective Laser Melting ................................. 14
   2.2 Preparation of Conventional Ti-6Al-4V fatigue samples ............................................... 16
   2.3 Characterization of microstructure .................................................................................... 16
   2.4 Investigation of surface roughness .................................................................................. 17
   2.5 Fatigue tests by 3-point bending ..................................................................................... 17
   2.6 Fracture surface examination .......................................................................................... 19

3. Results .................................................................................................................................. 20
   3.1 Microstructure of conventional Ti-6Al-4V samples .......................................................... 20
   3.2 Microstructure of Selective Laser Melting samples .......................................................... 22
   3.3 Surface roughness of conventional Ti-6Al-4V samples .................................................... 23
   3.4 Surface roughness of Selective Laser Melting Ti-6Al-4V samples .................................. 25
   3.5 Fatigue life of conventional samples ................................................................................. 27
   3.6 Fatigue life of Selective Laser Melting samples ............................................................... 28
   3.7 Evaluation of fracture surface Milled samples ................................................................. 29
      3.7.1 Conventional samples ............................................................................................... 29
3.7.2 Quenched samples ................................................................. 31
3.7.3 Quenched + Heat Treated samples ................................................... 32
3.8 Evaluation of fracture surface SLM samples ........................................... 33
4. Discussion .................................................................................... 35
  4.1 Effect of microstructure on fatigue life ............................................. 35
  4.2 Effect of surface roughness and morphology on fatigue life ............. 36
  4.3 Comparison of fractured surfaces .................................................. 38
5. Conclusions and recommendations .................................................... 41
6. Acknowledgements ....................................................................... 42
7. References .................................................................................... 43
1. Introduction

1.1 Applications of lightweight material

1.1.1 Needs

The need of lightweight materials becomes more essential in today’s society. For example, in the aerospace industry, implementing more lightweight materials would cut fuel consumption, increase speed and flight duration of the heavy aircrafts which would save both money and carbon emissions [1]. The growing demand for transportation of people and products puts pressure on material resources. When it comes to reducing energy consumption, researching in new lightweight material plays a crucial part, especially in the aerospace industry. In contrast to the automotive industry who are only willing to pay 10 Euros per one kilo of weight reduction, Leyens and Peters states that the aerospace industry would pay 1000 Euros for the equivalent weight reduction [2]. The difference between the industries is due to the fact that nearly one third of the aircrafts consist of fuel which would compare to a medium proportioned automotive carrying about 500 liters of fuel [2].

1.1.2 Conventional Titanium

The gains of lighter materials in aerospace applications are many and lightweight metals like aluminum and titanium alloys are already used in most parts of air crafts. As an example, aluminum alloys stand for approximately 80% of the airframe in an aircraft [2]. Due to its preferable properties, high specific strength, corrosion resistance, heat-resistance and compatibility to composites, titanium is used as the main substitute for steel where weight reduction is needed and secondly as space savings for aluminum [1]. Titanium ore, in contrast to other minerals, is evenly distributed around the world. This raises the question why titanium is not more widely used as its properties are preferred over current lightweight materials in many applications. The main reason is that production of titanium is far more expensive than other materials due to intricate extraction and melting procedures. A comparison made by Froes, states that sheets of titanium costs between 15-50 times more (dollars/pound) than aluminum [3]. This leads to the importance of developing new manufacturing processes for titanium and finding ways to lower material waste from conventional processes like milling and turning to make titanium economically beneficial compared to other lightweight materials. Furthermore, recent studies have focused on additive manufacturing
(AM) methods for its potential of replacing conventional manufacturing processes as products are manufactured straight out of metal powder to final shape without any material waste [1].

1.1.3 Conventional Titanium alloys

Similar to other metals titanium can occur in different crystal structures depending on alloy content and temperature. There are two main crystal structures relevant in the majority of Ti-alloys. Firstly, \( \alpha \) titanium, consisting of the hexagonal close packed (HCP) crystal structure, stabilizes at lower temperatures. Secondly, at elevated temperatures the (HCP) transforms into \( \beta \) titanium as body centered cubic (BCC) crystal structure. The differences of the two crystal structures leads to multiple ways of manipulating Ti-alloys and its mechanical properties. Furthermore, to understand how mechanical properties and microstructure varies with altered distribution of \( \alpha \) and \( \beta \), further knowledge of the two crystal structures are of concern. The (BCC) crystal consists of 12 independent slip systems in contrast to 3 of the (HCP) and this enables dislocations to move at a higher extent in (BCC) [2]. The freedom of dislocational movement is strongly connected to the ability to deform plastically and increase ductility whereas the compact and anisotropic planes of (HCP) provide good creep resistance [2].

The microstructure and transformation between \( \alpha \) and \( \beta \) can be controlled by adding alloying elements. Elements such as aluminum, oxygen, nitrogen and carbon are \( \alpha \) stabilizers meaning that adding of such element increase the stability of \( \alpha \) phase at elevated temperatures along with the creation of an \( \alpha + \beta \) region. Furthermore, the \( \beta \) transus can be lowered through the addition of \( \beta \) stabilizers. Elements like molybdenum, vanadium, niobium and tantalum make up a category of \( \beta \) stabilizers generating the \( \beta \) isomorphous phase diagram. The isomorphous stabilizers are essential due to their high solubility in the titanium matrix resulting in higher strength titanium alloys [2]. The influence of \( \alpha \) and \( \beta \) stabilizers on the titanium phase diagram is schematically illustrated in figure 1.1.
The composition between $\alpha$ and $\beta$ provides three main classes $\alpha$, $\alpha + \beta$, and $\beta$ titanium and it is interesting to compare how mechanical properties are altered within the classes. Inspired by Leyens and Peters, table 1.1 illustrates the influence of crystal structure on mechanical properties of titanium alloys with (+) meaning good and (-) bad [2]. This gives a good comprehension of how different fractions of alloying content manipulates the mechanical properties.

Table 1.1 Illustration of mechanical properties of $\alpha$, $\alpha + \beta$, and $\beta$ alloys. Inspired by [2].

<table>
<thead>
<tr>
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<th>$\alpha$</th>
<th>$\alpha + \beta$</th>
<th>$\beta$</th>
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<tr>
<td>Density</td>
<td>+</td>
<td>+</td>
<td>-</td>
</tr>
<tr>
<td>Strength</td>
<td>-</td>
<td>+</td>
<td>++</td>
</tr>
<tr>
<td>Ductility</td>
<td>-/+</td>
<td>+</td>
<td>+/-</td>
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<tr>
<td>Fracture toughness</td>
<td>+</td>
<td>-/+</td>
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<tr>
<td>Creep resistance</td>
<td>+</td>
<td>+/-</td>
<td>-</td>
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<tr>
<td>Corrosion behavior</td>
<td>++</td>
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<td>Oxidation behavior</td>
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<tr>
<td>Weldability</td>
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<tr>
<td>Cold formability</td>
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1.1.4 Influence of microstructure on mechanical properties of Ti-6Al-4V alloys

The most popular and thoroughly researched titanium alloy is Ti-6Al-4V an α + β alloy. The addition of 6% weight aluminum and 4% weight vanadium leads to a good combination of mechanical properties and the microstructure can be controlled by various heat treatments for specific applications. Adding aluminum makes titanium less dense while vanadium adds strength and temperature stability to the alloy [2]. According to Shunmugavel et al., the microstructure is of great importance of controlling properties of conventional titanium alloys. The composition of Ti-6Al-4V introduce a martensitic starting temperature and depending on cooling rates and the starting temperature of cooling, different microstructures are achievable [4]. Slow cooling from above the β transus results in a lamellar microstructure where α phase, upon entering the α + β zone, starts growing from β grain boundaries and fill up the prior β grains [2]. These prior β grains will consist of α lamella with a preferred orientation as the nucleation starts in one of the 12 earlier mentioned slip systems of the (BCC) crystal structure. Therefore, the orientation of the lamellar grains varies in relation to each other and do not act anisotropically. Furthermore, the cooling rate decides the size of the lamella where slow cooling leads to more diffusion resulting in coarse lamella. In contrary, increased cooling results in fine lamella because of less diffusion. Additionally, mechanical properties are directly governed by the size of the lamella with fine lamella indicating higher strength, ductility and resistance to fatigue crack nucleation compared to coarse lamella. On the other hand, coarse lamella shows increased fracture toughness, creep strength as well as resistance to fatigue crack propagation [2]. Quenching from above the β transus will form a needle like structure called α′. However, the martensitic structure differs from the brittle martensite observed in steels and Leyens and Peters states that α′ do increase strength of Ti-6Al-4V but the inner stresses of the crystal structure is not as high as in quenched steels, leading to a small increase in strength [2]. An equiaxed microstructure can be achieved by cold work acting as nucleation sites followed by recrystallization and solution treatment. In addition, equiaxed microstructure can also be divided in to fine and coarse microstructure depending on the time of annealing. Comparing mechanical properties, the equiaxed microstructure increases ductility, strength and exhibits better resistance to fatigue crack initiation than the lamellar microstructure. However, the equiaxed structure suffers a weakening in fracture toughness and creep strength in contrast to the lamellar structure, but solution heat treatment under the β transus leads to a combination of both equiaxed and lamellar microstructure called bimodular. The bimodular microstructure offers a good mixture of mechanical properties [2].
1.2 Additive manufacturing

Additive manufacturing (AM) is a relatively new manufacturing process, developed through rapid prototyping, which enables the making of lightweight materials with complex geometries compared to conventional methods. There are many different AM methods but most of them are using layer-by-layer building to achieve the final product. The possibilities are many and one of them is that AM uses raw material such as powder or wire which is locally melted by an energy source without any waste of material. The energy source, often in the form of a laser, is navigated by a computer-aided-design (CAD)-file. This leads to total freedom of designing AM products, which makes it possible to reduce weight of complex parts where conventional methods like turning need specialized tools or simply cannot access. Additionally, as earlier stated weight reduction is of great concern in the aerospace industry and AM offers the potential of reducing weight which would cut fuel consumption and thereby lower carbon emission and the cost of airplane travel [2,3]. Furthermore, AM has caught the eye from the biomedical industry because of the opportunity to fabricate individually specialized products like prosthetics. Here a lot of research has focused on Ti-6Al-4V for its light weight, compatibility to human bone modulus and compatibility to body environments [1-3]. However, implementing AM materials in structural applications such as the aerospace industry demands understanding of AM mechanical behavior. In contrary to conventional manufacturing processes, AM materials are not as thoroughly studied and therefore more research has to be done. The mechanical properties also depend on the AM method being used where the type of power source, thickness of building layers and other parameters governs the microstructure and mechanical properties of the end product [1]. The next section will cover the main types of AM methods used for manufacturing Ti-6Al-4V.

1.2.1 Classification of additive manufacturing methods

Additive manufacturing processes can be divided into different categories depending on raw material, the feeding strategy and the power source bonding the material. Furthermore, different approaches are suitable for different applications and a short introduction to the most common AM methods are presented.

Binder Jetting (BJ) offers a mixture of binder and powder, leading to a vast opportunity in mixing different materials. The powder consists of either metal or ceramic particles and make up the main material of the final product. On the other hand, the binder often an adhesive liquid sticks the building layers together [5]. Contrary to other AM processes BJ products are not done following the printing process, hence several post treatments have to be done like sintering leading to high
production costs. Additionally, the sintering process in combination with the liquid-based binder might result in porosities in between layers not making mechanical properties suitable for the demands of structures carrying loads [5]. Another developed AM method is Sheet Lamination (SL) where sheets of metal are stacked upon each other layer by layer resulting in the final product. Every layer is either pre-cut before bonding with the next layer or cut in a post bonding procedure. Furthermore, the bonding process of the layers can be done with adhesives as for BJ processes but also by laser welding and other metallurgical bonding processes [6]. In addition, SL main qualities are that the sheets retain their dimensions without getting distorted and also exhibits a rather fine surface to the end part [6]. As for BJ processes the use of adhesives may lead to problems implementing SL products in critical structures due to anisotropic characteristics in the building direction [6]. The most popular powder-based AM processes for metal products can be put in two major groups with the first being Powder Bed Fusion (PBF) where a power source locally melts areas on a powder bed and the powder is provided by a feeder filling the building platform for each layer. Secondly, Directed Energy Deposition (DED), also called Laser Cladding (LC), provides a different feeding strategy where powder is distributed locally and directly melted upon insertion. PBF include subgroups of methods based on the energy source and the most eminent ones are Selective Laser Melting (SLM) and Electron Beam Melting (EBM). In contrast to SLM, EBM uses an electron beam as a power source [7]. Several studies done by Li et al., have shown that the deep penetration of the electron beam results in thicker building layers decreasing build time, higher temperature in the chamber decreasing residual stresses and bigger grain size of powder lowering surface inoculants [3]. EBM is performed in vacuum which lowers the risk of unwanted light particles contaminating the product [1]. Additionally, the increased cooling rate has been observed to cause more internal stresses of SLM but generally the surface is less rough than products produced by EBM [1].

1.3 Selective Laser Melting

Selective Laser Melting, schematically illustrated in figure 1.2, is an AM method consisting of three general steps repeated for each building layer. In the first step an adjustable building-platform is lowered vertically, secondly raw material in form of metal powder is evenly distributed by a feeder. The final part consists of local melting by a laser which operates with the help of a CAD-file enabling layers between 30 and 150 micrometers thick [1,8]. SLM is performed in an inert gas environment to avoid oxidation and the process is conducted at elevated temperature reaching 200°C to lower the risk of impurities and residual stresses within the product [8]. Upon completion,
the product is taken out of the powder bed and unmelted powder can be recycled at a higher extent compared to EBM due to the laser not effecting surrounding powder not being a part of the built product as much. Thus, SLM is more sustainable than EBM for its easy recycling of unmelted powder. [8]

1.3.1 Advantages of Selective Laser Melting

Due to the precision of the laser, the ability of very thin layers can be achieved enabling manufacturing of complex geometries compared to traditional material removing and other AM methods. In addition, several studies imply that SLM can use smaller powder sizes ranging from 25 to 45 micrometers. In contrast, similar research performed on EBM used powder sizes altering between 45 and 300 micrometers [10]. Furthermore, the combination of these parameters makes SLM desirable, showing potential for several applications in important industries such as biomedical and aerospace [1-3, 8]. The earlier mentioned Ti-6Al-4V alloy, has been the object of most studies on SLM for its preferable properties such as biocompatibility, corrosion resistance and specific strength [1-3]. For example, SLM with its thin layer accuracy has shown interest by the biomedical industry with the potential of developing small complex cellular structures that could decrease density in an optimized fashion for biocompatible materials like Ti-6Al-4V. Additionally, a lower and more controlled density through cellular optimization could make prosthetics made of Ti-6Al-4V alloys comparable to the human bone and SLM has shown the potential [11,12]. The geometrical freedom and accuracy of SLM can as well reduce the material spill of this rather
expensive alloy with less manufacturing steps and lower environmental impact. Furthermore, lowering costs can make Ti-6Al-4V alloys affordable to replace materials with less wanted properties [1-4].

1.3.2 Disadvantages of Selective Laser Melting

SLM have many advantages as mentioned above. However, examinations of Ti-6Al-4V parts manufactured by SLM indicate multiple defects, reaching a size of 200 micrometers, and the defects were found to be related to unmelted powder and segregation of different phases in areas of the parts [12,13]. Because of high energy concentration by the laser followed by rapid cooling, the previous building layer exhibits large heat gradients in the vertical direction. This results in a microstructure with elongated grains of needle like martensite in the building direction [13]. In contrast to the vertical plane, elongated grains are not observed in the x and y plane of each layer but rather fine grains with randomly oriented packets of martensite. Thus, as-built samples show anisotropic properties [13].

As a consequence of the rapid cooling and high temperature variations during building, parts have shown the development of residual stresses. The residual stresses are particularly increased in the bottom and top layer of the parts where the greatest temperature differences occur. When a layer starts to solidify it will deform because of thermal contraction and the contraction is constrained by the underlying layer already being solidified. This phenomenon forms tensile stresses in the top layer and compression stresses in the previous layer [14]. The development of residual stresses during solidification are of great concern and a better understanding of process parameters has to be gained before implementing SLM parts for structural use. However, residual stresses have been shown to decrease by heating the building platform during manufacturing [14].

Furthermore, little is known about how process parameters influence the porosity. A study on how scanning speed influence porosity showed that the porosities increased at the lowest scanning speed. This is contradicting as a slower scanning speed increases the energy concentration giving the powder more energy to bond and densify [10]. In addition, the uncertainty in controlling porosity shows how intricate SLM is, making it hard to control mechanical properties. SLM suffers from constraints regarding the surface roughness which is based on several problems related to the building process. Firstly, SLM and other AM methods are effected by the Staircase Effect which occur through the deviation angle created between layers and it limits the geometrical freedom of SLM [15]. The roughness created by the staircase effect could be reduced by thinner layers. Thin layers have shown higher stabilization of the melt pool due to less powder included in every layer resulting in better surface roughness [16]. Controlling the surface roughness of the first
layer is of concern due to the fact that the initial scan is performed on pure powder and not on solidified metal. The combination of lower thermal conductivity and rapid energy development creates unstable melt pools and stalactite patterns are formed by the melt sinking through the powder by gravitation [16].

1.4 Influence of microstructure on mechanical properties of Selective Laser Melting Ti-6Al-4V alloys

As mentioned in previous sections, due to the rapid cooling of SLM, a martensitic microstructure is gained in the as-built condition of Ti-6Al-4V alloys similarly to quenched wrought Ti-6AL-4V alloys [1,8,12]. Additionally, it is of interest to know how the as-built microstructure of SLM Ti-6Al-4V influences the mechanical properties as the main purpose of AM methods is to produce net shape products without further treatments making it less sustainable than conventional methods [1,8,12]. Leuders et al., studied how tensile properties changed after different heat treatments of the as-built microstructure. The conclusion was that the as-built microstructure lost mechanical strength but increased ductility following further heat treatments at 800°C and 1050°C. In addition, the change in tensile properties was strongly connected to grain growth and higher detection of the β phase after the heat treatments [8]. The mechanical properties of the as-built structure was also examined by Gong et al., comparing tensile properties of Ti-6Al-4V produced by SLM and EBM where SLM samples proved greater tensile and yield strengths. The difference was contributed by the martensitic structure of the SLM whereas EBM was performed at higher temperatures leading to a fine lamellar microstructure of α and β. Thus, confirming the role of microstructure on mechanical properties [17].

1.5 Fatigue in general

Fatigue is the most common reason for failure of mechanical structures subjected to loads. Failure can occur at high static loads above the ultimate tensile strength (UTS) but also at loads significantly lower. However, fatigue is the result of cyclic loading where the applied stress alternates over time leading to a different fracture characteristic compared to static loads. Fatigue originates from high stress concentrations caused by material or geometrical defects acting as crack initiation sites. Small cracks start to develop, gradually growing for each loading cycle, until it reaches a critical size
causing final fracture. The initiation of a crack and its propagation can be detected by studying the fracture surface [18].

As mentioned, fatigue is influenced by different types of loads, acting on a structure, varying over time. An aircraft is a perfect example where temperature, forces and other loads varies during flights. Thus, one of the most common reasons for failures in the aerospace industry is due to fatigue caused by cyclic loading [19].

To predict fatigue life, cyclic stresses are modelled as sinusoidal functions, see figure 1.3, where the important parameters related to fatigue are illustrated. In addition, equation 1.5.1-1.5.3, shows the relationship between these parameters. Where the mean stress ($\sigma_{mean}$), equation 1.5.1, is defined as the mean value between the maximum and minimum stress. The amplitude stress ($\sigma_{amp}$), equation 1.5.2, is the mean difference between maximum and minimum stress [18,19]. The combination and altering of the stress parameters enables different cyclic loading scenarios, which is convenient as mechanical components can be subjected to very peculiar loading conditions. To obtain as good approximations of a component’s fatigue behavior as possible, the testing method should be comparable to the component’s type of load in its application. As an illustration, when a structure is predicted to be subjected to a bending stress, accordingly its fatigue characteristics should be studied in a bending machine [19]. However, an important parameter for understanding fatigue behavior is the stress ratio (R), see equation 1.5.3. The stress ratio is defined as the ratio between the minimum stress and the maximum stress subjected to a mechanical component throughout an entire loading cycle. These values could either be negative or positive where a negative ratio symbolizes a compression stress and positive ratio a tensile stress. The common used stress ratios are, R=0, R=-1 and R=0.5 where R=0 is defined as a stress alternating between zero and the maximum stress. This is referred to as Unidirectional Stress whereas the more severe loading condition, R= -1, is called Fully Reversed Stress characterized by the mean stress equaling zero and the same value of maximum and minimum stress in different directions. Furthermore, a Fluctuating Stress could be used, as for R=0.5, where a preload is applied and the sample is always under influence of a certain load [18,19].
Figure 1.3 Schematic of cyclic stress ($\sigma(t)$) as sinusoidal function of time ($t$) where the amplitude stress ($\sigma_{ampl}$), mean stress ($\sigma_{mean}$), maximum stress ($\sigma_{max}$) and minimum stress ($\sigma_{min}$) are presented

\[
\sigma_{mean} = \frac{\sigma_{max} + \sigma_{min}}{2} \\
\sigma_{ampl} = \frac{\sigma_{max} - \sigma_{min}}{2} \\
R = \frac{\sigma_{min}}{\sigma_{max}}
\]  

(1.5.1)  
(1.5.2)  
(1.5.3)

1.5.1 3-Point bending

As stated above various test methods for characterization of fatigue behavior have been developed and a method specifically made for mimicking components under the influence of flexural stresses is 3-point bending. The equipment consists of two vertically movable support pins holding the sample along with a pin located on top of the sample applying the load. When a load is applied the sample will experience a bending moment starting from zero at the support pins reaching the highest value at the center where the applied stress is located. Thus, the test specimen will experience a compressive stress at the top of the specimen and a tensile stress at the bottom surface.
For materials with rough surfaces the fracture would likely occur on the tension side of a sample because of the increased stress concentrations at the rough surface leading to crack initiation [20-21].

1.5.2 Fatigue of Selective Laser Melting Ti-6Al-4V alloys

Due to its short history, several studies on SLM Ti-6Al-4V alloys have focused on fatigue behavior [1,4,7,8]. Studies made by Leuders et al., investigated how fatigue performance was influenced by heat treatment. The fatigue life of SLM Ti-6Al-4V was shown strongly connected to defect and microstructural behavior when the as-built martensitic condition went through different heat treatments. Two hot isostatic pressure heat treatments (HIP) were performed at 1050°C and 920°C respectively. Compared to heat treatments without HIP treatments it was found that HIP managed to increase fatigue life, mainly because of decreased porosity. However, a lower fatigue life was observed by the 1050°C heat treated conditions compared to the 920°C samples. This was found to be influenced by the lowered strength as some α phase was transformed to the more ductile β phase, more susceptible for local deformation resulting in crack initiations [8].

Furthermore, as mentioned in previous sections the building procedure of SLM have shown to develop anisotropic behavior due to elongated grains in the building direction [1,13,23]. Bending fatigue was performed by Nicoletto, investigating how anisotropic properties influenced the fatigue life of samples built in different orientations, see figure 1.4. In addition, it was shown that samples built in the z-x direction had inferior fatigue life due to the applied force being perpendicular to elongated grains. Thus, anisotropic behavior of SLM products have to be considered regarding the fatigue life of components in different applications [1,23].

The fatigue behavior was also found to be influenced by the poor surface roughness contributed by the SLM building process [1,24]. The surface roughness was stated to be the biggest contribution to the low fatigue life of as-built surfaces and that machining of the surface substantially increased fatigue life. The top layer morphology indicated sharp geometrical deviations acting as small notches leading to high stress concentrations being susceptible to crack initiation.
1.6 Aims

Cheap and sustainable manufacturing processes for Ti-6Al-4V parts are needed to meet future global demand. Selective Laser Melting is an additive manufacturing method with the potential to fulfill those demands. The ability to produce complex products directly from metal powder without the multiple steps of conventional manufacturing would decrease the price of implementing Ti-6Al-4V in its different applications. However, as SLM is a relatively new manufacturing process there is a lack of knowledge and further understanding of how mechanical properties are affected has to be done. Therefore, this thesis aims to investigate how microstructure, surface roughness and the morphology of top surface influence fatigue life by 3-point bending of Ti-6Al-4V samples made by SLM. Furthermore, the fatigue life of conventional manufactured Ti-6Al-4V specimens are analyzed under similar conditions as a reference for understanding SLM.
2. Method

The following section covers materials and methods used for achieving the goal of this project. The materials used for this thesis are all consisting of the Ti-6Al-4V alloy. Both SLM and conventional samples were prepared through different heat treatments, investigated for characterization followed by fatigue tests.

2.1 Preparation of fatigue samples made by Selective Laser Melting

Ti-6Al-4V alloy feedstock powder was obtained from TLS Technik GmbH (Germany). The powder had a spherical morphology and a particle size distribution of 10 percent below the size of 7.6 µm (Ø10=7.6 µm), 50 percent below 28.6 µm (Ø50=28.6 µm) and 90 percent of the particles below 41.9 µm (Ø90=41.9 µm). The SLM equipment used was a ProX® DMP 200. The building chamber was filled with argon gas to prevent oxidation. 12 samples were produced on the same construction plate and had the same scan strategy and parameters giving optimum sample density. The fatigue specimens were dissociated from the plate by wire Electrical Discharge Machining (EDM) after heat treatment 2h at 940°C followed by furnace cooling with Argon and 2h at 650°C.

The samples were milled on 5 out of 6 faces, using the same milling parameters in order to calibrate the specimen dimensions (8 mm x 8 mm x 60 mm). The surface to test was the only surface kept untouched. In order to obtain 3 different surface morphologies, relative to the length of the samples (30°, 60° and 90°), the laser scan direction was altered in 3 ways for the last layer. Out of the 12 samples, 4 were given the same of the 3 orientation angles. Figure 2.1 illustrates the last layer scan strategy and the following conditions were produced:

1. γ = 90° (θ = 85°)
2. γ = 60° (θ = 55°)
3. γ = 30° (θ = 25°)
Figure 2.1 Schematic of the SLM method and the laser scanning strategy of the top layer relative to the x-y axis of the building platform.

Figure 2.2 illustrates the 3 different surface morphologies gained by altering the laser scanning direction relative to the length of the samples. The red lines indicate the path of the laser and leads to a surface morphology throughout the entire surface with a certain angle ($\gamma$) relative to the length of the sample.

Figure 2.2 Schematic of laser scanning direction relative to the length of the sample for obtaining 3 different surface morphologies (30°, 60° and 90°) and the relation to x-y axis of the building platform.
2.2 Preparation of Conventional Ti-6Al-4V fatigue samples

Titanium ASTM F136 grade 5 (Ti-6Al-4V) was bought as round bars from Harald Phil AB in a standard mill annealed condition and then cut into 18 different samples of the same rectangular dimensions as illustrated in figure 2.3. Following the cutting, two different heat treatments were performed to obtain three different microstructures. 12 of the samples were heat treated in a furnace at 1050°C for 20 minutes followed by quenching in water. Of the 12 quenched specimens 6 were further heat treated in the same way as the SLM samples to obtain comparable microstructures. This resulted in three different conditions to be examined the conventional (C), the quenched (Q) and the quenched + heat treated (Q+HT).

To achieve a similar surface roughness and morphology of the top layer as the SLM specimens, the conventional samples were milled. The milling pattern was conducted in three orientations, relative to the length of the sample to obtain surface morphologies of (30°, 60° and 90°). Furthermore, two samples of each condition and morphology were made. All samples are illustrated in table 2.

Table 2.1 Conventional Ti-6Al-4V samples to investigate.

<table>
<thead>
<tr>
<th>Condition</th>
<th>30°</th>
<th>60°</th>
<th>90°</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>x2</td>
<td>x2</td>
<td>x2</td>
</tr>
<tr>
<td>Q</td>
<td>x2</td>
<td>x2</td>
<td>x2</td>
</tr>
<tr>
<td>Q+HT</td>
<td>x2</td>
<td>x2</td>
<td>x2</td>
</tr>
</tbody>
</table>

2.3 Characterization of microstructure

Characterization of the microstructure was done by an Optical Microscope Leica DMI3000 M obtaining images at three different magnifications (x10, x20 and x50) for each condition (C, Q and Q+HT). Prior the investigation, the samples were grinded in several steps using different grit sizes.
First using PIANO 120 silicon carbide paper and secondly, 9 micrometer Allegro diamond particles. After grinding the samples were polished using fine 3 micrometer Allegro. These processes were conducted to remove scratches and dirt and etching was also done using Kroll’s reagent for 30 seconds to be able to visualize the microstructure in the microscope. The same strategy was used for the SLM sample but in contrast to the Conventional samples two cross sections were examined of the SLM sample. One in the building direction (z-x plane) and another cross section in the x-y plane.

2.4 Investigation of surface roughness

The surface roughness of all the specimens were investigated by a Bruker ContourGT-I 3D profilometer. The profilometer was used to scan a small area of 5x5 mm in the central part of the specimens to obtain the mean height of the surface roughness ($S_a$), the maximum valley and maximum peak summarized ($S_z$), the maximum peak value ($S_p$) and the maximum valley ($S_v$) [25]. These surface parameters were used to compare the surface roughness of SLM and Conventional samples. By using the software Vision64, 2D and 3D images were obtained illustrating the roughness as well as the surface morphologies. The procedures were made for both the milled Conventional samples and the SLM samples.

2.5 Fatigue tests by 3-point bending

Fatigue tests were performed on an Instron 3-point bending machine. The cyclic loading was kept at a constant stress ratio $R = 0.1$, see section 1.5. In Table 2.2 the calculated stresses are presented where the maximum stress 700 MPa was chosen in accordance to previous fatigue data of SLM Ti-6Al-4V gathered by Li et al., and consideration was taken to the time frame of this project where fatigue life was wanted in the region of $10^4$ to $10^6$ cycles [12]. The software running the Instron was working with loads in Newton (N) as input data and accordingly, the stress parameters (mean and amplitude) had to be converted to forces using equation 2.5.1 [21]. In Figure 2.4, a schematic of the 3-point bending machine is illustrated along with the relevant geometric parameters used to calculate the input data. To be certain that the same cyclic load was applied for all samples the width (a) and thickness (b) of equation 2.5.1 was measured for each sample using a calipers. The fatigue tests were set to run for $10^6$ cycles and the specimens that did not fail were considered run outs. Samples were put with the prepared surface resting on the support pins. Thus, making the 90° morphology perpendicular to the applied load. As an example, figure 2.5 illustrates how the
applied load relates to the 90° surface morphology relative to the length of the sample, during the 3-point bending test. The image also shows that the surface of interest is subjected to tension during the tests [21].

Table 2.2 Stress parameters for cyclic loading.

<table>
<thead>
<tr>
<th>LABEL</th>
<th>STRESS</th>
</tr>
</thead>
<tbody>
<tr>
<td>σMAX</td>
<td>700 MPa</td>
</tr>
<tr>
<td>σMIN</td>
<td>70 MPa</td>
</tr>
<tr>
<td>σMEAN</td>
<td>385 MPa</td>
</tr>
<tr>
<td>σAMP</td>
<td>315 MPa</td>
</tr>
</tbody>
</table>

\[
\sigma = \frac{3FL}{2\alpha b^2} \quad \Leftrightarrow \quad F = \frac{2\alpha b^2 \sigma}{3L} \quad (2.5.1)
\]

Figure 2.4 Schematic of a 3-point bending test and geometrical parameters.
Figure 2.5 Schematic of $\gamma=90^\circ$ sample placed in the fatigue test and how the surface morphology correlates to the applied stress and support beams of the 3-point bending test.

2.6 Fracture surface examination

Following the fatigue tests the fracture surfaces of the broken samples were examined by a LEO 1530 Scanning Electron Microscope (SEM). The SEM used an electron beam with an energy of 20 (kV) and it was performed in vacuum. The goal was to localize the crack initiation of each specimen along with the detection of porosities and apparent defects. The samples were also tilted at different angles to get a better comprehension of the crack path and its relation to the surface morphology.
3. Results

The following section covers the results of microstructure characterization by Optical microscope, surface evaluation using profilometer, fatigue life by 3-point bending tests and evaluation of fracture surfaces by Scanning Electron Microscope.

3.1 Microstructure of conventional Ti-6Al-4V samples

The following microstructures were obtained by optical microscope after etching to enhance the α and β phases of the different heat treated Ti-6Al-4V samples. The conventional sample, shown in figure 3.1, revealed a fine microstructure consisting of equiaxed α and globular β grains. This is a standard for the as delivered Ti-6Al-4V after mill annealing.

Following the heat treatment at 1050°C for 20 minutes and rapid cooling in water, the quenched sample showed a significant grain growth compared to the conventional sample. The new prior β grains contained acicular α’ martensite as a consequence of the rapid cooling. In figure 3.2 one can
clearly see the needle like dark $\alpha'$ with the same orientation, within every grain, but different in relation to other grains.

The characterization of the last sample, illustrated in figure 3.3, which went through further heat treatment after quenching, indicated a small increase in grain growth and the packets of $\alpha'$ transformed into thin lamellar platelets of $\alpha$ along with $\alpha$ phase developed at the $\beta$ grain boundaries. All three conditions are illustrated in figure 3.4 at a higher magnification indicating both grain growth from left to right and an increased lamellar size from Q to Q+HT.
3.2 Microstructure of Selective Laser Melting samples

Optical micrographs of the Ti-6Al-4V samples manufactured by SLM at low to high magnification along the x and z direction is illustrated in figure 3.5. The microstructure is characterized by
multiple elongated grains in the building direction (Z). These grains contain similar oriented packets of α lamella but varying in relation to the other elongated grains. Images were taken from a cross section in the x and y direction showing no elongated grains but fine, square like grains approximately 100 micrometers wide. The different textures of the two cross sections examined imply anisotropic properties. A comparison between the two cross sections is illustrated in figure 3.6.

Figure 3.5. Images of x-z direction (SLM Ti-6Al-4V heat treated) consisting of elongated grains with packets of alpha with magnifications 10x, 20x and 50x.

Figure 3.6. Microstructure of x-y cross section (a) and x-z cross section (b).

3.3 Surface roughness of conventional Ti-6Al-4V samples

The following results contain the topology characteristics of the milled conventional samples gained by using a Profilometer. The surface roughness parameters, illustrated in table 3.1, were calculated using the software Vision64. 2D and 3D images for visualization of the different surface morphologies are represented in figure 3.7-3.9. The results show that a similar roughness was achieved with a mean value of \( S_a = 5.823 \mu m \) for the different milled orientations investigated. Looking at figure 3.7-3.10 it is clear that an approximate angle of 90°, 60° and 30° was obtained by
the milling. However, constant angles over the entire x-y plane of the specimens were not achievable using milling.

Table 3.1 Surface parameters milled specimens with different orientations.

<table>
<thead>
<tr>
<th>Milled Orientation</th>
<th>90°</th>
<th>60°</th>
<th>30°</th>
<th>Mean</th>
</tr>
</thead>
<tbody>
<tr>
<td>$S_a$ ($\mu m$)</td>
<td>5.904</td>
<td>5.595</td>
<td>5.969</td>
<td>5.823</td>
</tr>
<tr>
<td>$S_z$ ($\mu m$)</td>
<td>41.355</td>
<td>32.220</td>
<td>39.151</td>
<td>37.575</td>
</tr>
<tr>
<td>$S_p$ ($\mu m$)</td>
<td>24.755</td>
<td>17.102</td>
<td>21.738</td>
<td>21.198</td>
</tr>
<tr>
<td>$S_v$ ($\mu m$)</td>
<td>-16.600</td>
<td>-15.118</td>
<td>-17.413</td>
<td>-16.377</td>
</tr>
</tbody>
</table>

Figure 3.7 Images of the 90° milled surface orientation, 2D and 3D.

Figure 3.8 Images of the 60° milled surface orientation, 2D and 3D.
3.4 Surface roughness of Selective Laser Melting Ti-6Al-4V samples

The surface parameters of the SLM samples are presented in table 3.2 along with 2D and 3D images illustrated in figure 3.10-3.12. The results were obtained in the same way as the conventional samples. The mean height of the surfaces (\( S_a \)) was calculated to 9.267 \( \mu \text{m} \) and by looking at figure 10-12 it's clear that the different orientation of 90°, 60° and 30° were obtained with straight patterns not deviating from the angles. Furthermore, the images imply a vast amount of surface defects that could act as stress concentration during loading.

Table 3.2 Surface parameters SLM specimens with different orientations.

<table>
<thead>
<tr>
<th>Surface Orientation</th>
<th>90°</th>
<th>60°</th>
<th>30°</th>
<th>Mean</th>
</tr>
</thead>
<tbody>
<tr>
<td>( S_a ) (( \mu \text{m} ))</td>
<td>9.090</td>
<td>9.425</td>
<td>9.285</td>
<td>9.267</td>
</tr>
<tr>
<td>( S_z ) (( \mu \text{m} ))</td>
<td>66.441</td>
<td>107.65</td>
<td>86.834</td>
<td>86.975</td>
</tr>
<tr>
<td>( S_p ) (( \mu \text{m} ))</td>
<td>36.464</td>
<td>79.985</td>
<td>55.061</td>
<td>57.170</td>
</tr>
<tr>
<td>( S_v ) (( \mu \text{m} ))</td>
<td>-29.977</td>
<td>-27.666</td>
<td>-31.773</td>
<td>-29.805</td>
</tr>
</tbody>
</table>
Figure 3.10 Images of the 90° SLM surface orientation, 2D and 3D.

Figure 3.11 Images of the 60° SLM surface orientation, 2D and 3D.

Figure 3.12 Images of the 30° SLM surface orientation, 2D and 3D.
3.5 Fatigue life of conventional samples

The values presented in table 3.3, represents the number of cycles to failure of the conventional samples by 3-point bending test. The tests were running for $10^6$ cycles and the samples that did not break were considered run outs. The results are also illustrated in figure 3.13.

Table 3.3 Number of cycles to failure ($N_F$) of the Conventional samples.

<table>
<thead>
<tr>
<th>Label</th>
<th>$N_F$</th>
<th>Label</th>
<th>$N_F$</th>
<th>Label</th>
<th>$N_F$</th>
</tr>
</thead>
<tbody>
<tr>
<td>C-30-1</td>
<td>&gt;1000000</td>
<td>Q-30-1</td>
<td>&gt;1000000</td>
<td>Q+HT-30-1</td>
<td>&gt;1000000</td>
</tr>
<tr>
<td>C-30-2</td>
<td>301384</td>
<td>Q-30-2</td>
<td>462125</td>
<td>Q+HT-30-2</td>
<td>241066</td>
</tr>
<tr>
<td>C-60-1</td>
<td>60376</td>
<td>Q-60-1</td>
<td>&gt;1000000</td>
<td>Q+HT-60-1</td>
<td>87024</td>
</tr>
<tr>
<td>C-60-2</td>
<td>256282</td>
<td>Q-60-2</td>
<td>&gt;1000000</td>
<td>Q+HT-60-2</td>
<td>66844</td>
</tr>
<tr>
<td>C-90-1</td>
<td>90387</td>
<td>Q-90-1</td>
<td>332261</td>
<td>Q+HT-90-1</td>
<td>130702</td>
</tr>
<tr>
<td>C-90-2</td>
<td>74576</td>
<td>Q-90-2</td>
<td>119044</td>
<td>Q+HT-90-2</td>
<td>65452</td>
</tr>
</tbody>
</table>

Figure 3.13 Cycles to failure ($N_F$) as a function of surface morphology (deg).
3.6 Fatigue life of Selective Laser Melting samples

In figure 3.14 results from 3-point bending tests are represented for the SLM samples. The number of cycles to failure is plotted as a function of surface orientation. The graph indicates that the 60° samples had longer fatigue life than 30° and 90° specimens and the average cycles to failure for all orientations are gathered in table 3.4.

Table 3.4 Mean value of cycles to failure (N_t) for the SLM specimens.

<table>
<thead>
<tr>
<th>Sample</th>
<th>Mean N_F</th>
</tr>
</thead>
<tbody>
<tr>
<td>SLM-30</td>
<td>16761</td>
</tr>
<tr>
<td>SLM-60</td>
<td>20915</td>
</tr>
<tr>
<td>SLM-90</td>
<td>15314</td>
</tr>
</tbody>
</table>

Figure 3.14. Cycles to failure (N_t) as a function of surface morphology (deg) of the SLM samples.
3.7 Evaluation of fracture surface Milled samples

Post fatigue tests the fracture surfaces were studied in SEM trying to locate the crack initiation as well as the reason for failure. The results are illustrated for every condition, one at a time, regarding heat treatment and surface orientation.

3.7.1 Conventional samples

The C samples all broke except the C-30-1 specimen which was considered a run out. Sample C-30-2 managed the highest amount of cycles to failure (301384) and in figure 3.15, the upper right corner, the crack initiation can be visualized starting in the surface followed by propagation through multiple paths. Furthermore, both the C-60 samples indicated the same crack initiation, starting from the same corner as the C-30-2. In figure 3.16, defects from the milling are visible and can be assumed to act as starting points of the cracks giving a higher stress concentration. Figure 3.16 shows that the crack propagated perpendicular to loading direction. Figure 3.17 represents the C-60-1 and C-60-2 samples where different sizes and geometries of the milling defects were found at the crack initiations. In addition, the lower fatigue life of C-60-1 compared to C-60-2 could be the result of the sharp edge and the larger defect size, approximately 300 µm wide. In contrast the C-90-2 sample indicates a rounded edge and smaller defect size of 250 µm wide. As well as the other samples the crack initiation started at the surface due to milling defects for the C-90 samples.
Figure 3.15 Crack initiation of the C-30-2 sample, starting in a milling defect in the top right corner.

Figure 3.16 Tilted view of the crack initiation of the C-60-1 sample showing an approximate 300 µm milling defect at lower magnification to the left and higher to the right.
Figure 3.17 Fracture surfaces of C-60-1 to the left and C-60-2 to the right. The milling defects and crack initiations can be seen in the upper left corner of each sample.

3.7.2 Quenched samples

Both the Q-90 samples broke and the fracture surfaces of Q-90-1 and Q-90-2 are illustrated in figure 3.18, were crack initiation occurred in the top right corner. By looking at the tilted view of figure 3.19 (b), a large milling defect was found on the surface of the Q-90-2 sample. This defect could be assumed to cause the crack initiation. The rest of the Q samples did not fracture except the Q-30-2 specimen. However, due to the limited time frame the Q-30-2 sample was excluded from SEM evaluation.

Figure 3.18 SEM images of Q-90-1 (a) and Q-90-2 (b) crack initiation top right corner for both samples.
3.7.3 Quenched + Heat Treated samples

The Q+HT-60 and Q+HT-90 samples had similar crack initiations starting at the surface. Figure 3.20 (a), shows that the crack initiation of Q+HT-60-1 had the same crack location as found in previous sections. Accordingly, a defect contributed by the milling, seen in a tilted view in figure 3.20 (b), seems to have initiated the crack that propagated perpendicular to the applied cyclic load. In contrast to the other heat treated samples, Q+HT-30-2 did not fracture in the center where the highest load is located (see section 2.1.4). Figure 3.21 shows an assumed pore or defect located in the circled area of the image. This may be the cause of the Q+HT-30-2 sample not breaking at the center of the surface.
3.8 Evaluation of fracture surface SLM samples

In contrast to the milled samples, SLM fracture surfaces were investigated by an optical microscope. In figure 3.22 the fracture surfaces of all three surface orientations are illustrated accordingly, (a) 30°, (b) 60° and (c) 90°. Figure 3.22 reveals multiple crack initiations at the rough surface, seen in the top part of the samples. Looking at the profiles of the fracture surfaces, image (b) illustrating the 60° SLM sample indicates a less rough surface profile which could be the reason for higher fatigue life compared to the other orientation showing a similar roughness profile. The path of the cracks at different orientations can be seen in figure 3.23.
Figure 3.23 Image in tilted view of surfaces of SLM samples 30° (a), 60° (b) and 90° (c).
4. Discussion

The goal of this thesis was to investigate how microstructure, surface roughness and the morphology of the top surface influenced fatigue behavior of SLM Ti-6Al-4V alloys by 3-point bending tests. Furthermore, conventional Ti-6Al-4V was tested as a reference to understand the differences in fatigue behavior. This section analyses the results and methods used and a comparison with previous studies will be covered.

The results of the 3-point bending test showed that the SLM Ti-6Al-4V samples had a significantly shorter fatigue life than the conventional samples. Taking all the conventional samples in consideration, the mean number of cycles to failure of the SLM samples was only 9% of the fatigue life measured of the conventional specimens. To understand these results the different parameters investigated need to be considered.

4.1 Effect of microstructure on fatigue life

The microstructural characterization showed a significant difference between the heat treated SLM Ti-6Al-4V samples and the conventional samples subjected to the same heat treatment, see figure 4.1. In accordance with previous studies, the SLM samples showed elongated prior $\beta$ grains containing $\alpha$ lamella reaching up to 1 mm in the building direction and a finer equiaxed lamellar structure in the x-y direction indicating anisotropic properties [1,13,23]. In contrast, the heat treated conventional microstructure contained larger grains and thicker lamellas assumed to be equally distributed throughout the microstructure. Leyens and Peters, discussed the influence of microstructure related to crack initiation, see section 1.1.4, where a fine lamella of Ti-6Al-4V alloys in contrast to a coarse was claimed to be more resistant to crack nucleation due to higher ductility and strength [2]. This was not seen in the results of the 3-point bending tests as SLM samples showed lower fatigue life than conventional Q+HT specimens. Even though inherent pores were not found in SLM samples it cannot be ruled out, Leuders et al., discussed how the heat treatment at 1050°C could increase the presence of the more ductile $\beta$ phase close to inherent pores of SLM samples. In addition, this could have resulted in local deformation in such regions decreasing fatigue life of SLM samples whereas conventional Ti-6Al-4V do not show the presence of inherent defects in the literature [8]. Between all the samples investigated the fatigue life ranked from highest to lowest were Q, C, Q+HT and SLM samples. This indicates that the higher strength of the martensitic microstructure could be the reason of longer fatigue life.
Figure 4.1 Optical micrographs illustrating the differences between conventional microstructure (a), and SLM samples from two different cross sections (b) and (c) after the same heat treatment.

4.2 Effect of surface roughness and morphology on fatigue life

The surface roughness and the morphology of the surface was clearly seen from the results of the profilometer. The mean arithmetic height $S_a$ had a significantly higher value for the SLM samples ($S_a \approx 9.3 \mu m$), compared to the conventional specimens ($S_a \approx 5.8 \mu m$). This is believed to be the main reason for the lower fatigue life of SLM samples. Looking at the 3D morphology of the 90-degree conventional and SLM sample, seen in figure 4.2 (a) and (b), it is obvious that the milling of the conventional samples did not result in the same surface characteristics as the SLM samples. Firstly, the 90-degree morphology is not perfect for the conventional sample, the milling strategy resulted in smooth deviations from the angle not making the pattern perfectly aligned with the applied load which could mean less stress concentrations in comparison to the straight pattern of the SLM sample, resulting in longer fatigue life. Secondly, the 4.2 images reveal that the surface of the SLM sample is rougher in every direction and shows a greater variety of peaks and valleys than the conventional sample. In contrast, the conventional sample is only rough in one surface profile. All these factors might have contributed to earlier and multiple crack initiations of the SLM samples, as more irregularities of uncontrolled dimensions acts as stress concentration notches, as seen in literature [1,24]. This difference was evident for all the different morphologies investigated, see section 3.3 and 3.4.

The mean logarithmic cycles to failure as a function of surface morphology is illustrated in figure 4.3, for both SLM and conventional samples. Regarding the morphology, the 90$^\circ$ angle perpendicular to the applied stress, illustrated the worst fatigue performance taking all samples in consideration. None of the 90$^\circ$ samples survived the full $10^6$ cycles as some of the 60$^\circ$ and 30$^\circ$
samples did. Thus, indicating that a surface morphology perpendicular to the applied force decreases fatigue life regardless of microstructure and that a decrease of that angle could increase fatigue life. However, finding the best morphology was challenging as different heat treatments showed varying trends. The C samples, as seen in figure 4.2, showed almost a linear increase in fatigue life by shifting the pattern of the surface from $90^\circ$ to $30^\circ$. In contrast, the SLM samples had the longer fatigue life at the $60^\circ$ morphology and the differences between $90^\circ$ and $30^\circ$ were minor. It could be assumed that the rough surface seen in the profilometer made the SLM samples less influenced by the morphology of the surface as random defects from manufacturing could have acted as notches giving higher stress concentrations in certain areas of the surface leading to multiple crack initiations [24].

Figure 4.2 3D images illustrating how different the surface roughness and morphology of the conventional $90^\circ$ sample (a) is compared to the SLM $90^\circ$ (b).
4.3 Comparison of fractured surfaces

To understand how the microstructure, surface roughness and morphology of the surface influenced fatigue life the origin of crack initiation should be compared between SLM and conventional samples.

Most of the conventional specimens had crack initiations in the same corner of the top surface as can be seen in figure 4.4, illustrating all the conventional conditions as follows, C (a), Q (b) and Q+HT (c). In addition, the circled areas show that crack initiation took place in defects created during the milling process. The number of cycles to failure could be connected to the size and geometrical features of the milling defects as illustrated in figure 4.5 (a) and (b). The C-60-1 (a), failed earlier than C-60-2 (b), reaching a fatigue life of $N_f=66844$ and $N_f=256282$ respectively. By looking at the separate milling defects, the shorter fatigue life of C-60-1 is assumed to be caused by the sample showing a larger and sharper milling defect which leads to higher stress concentrations.
 Compared to the milled samples the SLM specimens did not break in the center of the applied load, but broke randomly at a small distance towards one of the two support pins. Even though the crack initiation was not as clearly visualized, as for the conventional samples, the images of the fracture surfaces indicate multiple crack initiations at the rough surface. In figure 4.6, the fracture surfaces of both SLM and Q+HT conventional samples are illustrated. By observing the SLM sample a) of figure 4.6, a large amount of irregularities are visible along the entire surface profile and these are of different sizes and geometries as seen of the depth and sharpness of the groove pointed out in the image. These grooves have been shown to act as stress concentrations leading to multiple crack nucleation sites [24]. The poor surface quality is connected to the manufacturing process of SLM where the rapid cooling prevents the powder to solidify homogenously giving a rough surface [12,13,24]. In contrast, the heat treated conventional sample of figure 4.6 (b), has a smoother surface profile looking passed the milling defect. Thus, the surface roughness of the SLM samples is assumed to be the most critical parameter leading to only 9% of the fatigue life of conventional samples.
Figure 4.6 Comparison of fracture surfaces between a SLM sample (a) and (b) a \(Q+HT\) conventional sample indicating different surface profiles.
5. Conclusions and recommendations

By investigating how microstructure, surface roughness and the morphology of the surface influenced the fatigue performance of SLM Ti-6Al-4V alloys and by conducting the same examinations on conventional Ti-6Al-4V alloys as a reference for better knowledge. The following conclusions could be made.

- The microstructure of the SLM Ti-6Al-4V samples are significantly different from conventional Ti-6Al-4V. The SLM process leads to anisotropic properties compared to conventional microstructure and this should be taken in consideration when applying SLM made products in load carrying structures.

- The SLM Ti-6Al-4V samples only managed 9% of the fatigue life of the investigated conventional samples mainly because of the poor surface roughness of $S_a \approx 9.3 \mu m$ compared to $S_a \approx 5.8 \mu m$ of the conventional samples.

- The 90° surface morphology proved to have the worst fatigue life of all the investigated Ti-6Al-4V samples because of the pattern being perpendicular to the applied stress. By shifting the surface pattern to 60° and 30° higher fatigue life could be obtained. However, this morphology dependence was not as clear for SLM samples as for conventional samples indicating that further research should be done.

- All the conventional samples failed due to milling defects located on the corner of the surface. The method of milling the conventional samples to gain a similar surface roughness and morphology as the SLM samples was not optimal and future work should be guided towards using a different strategy if wanting to get a similar surface as SLM samples.

- The SLM samples did not break in the center of the highest load probably because of randomly placed surface defects acting as stress concentrations. This could be a result of the manufacturing process where regions of unmelted powder and lack of fusion could have contributed.
6. Acknowledgements

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7. References


