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Microstructure evolution of 316L produced by HP-SLM (high power selective laser melting)

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Abstract:

New generation of selective Laser Melting (SLM) machines are evolving towards higher power lasers as well as multi laser systems in order to increase the productivity. The increase in laser power and the modification of the laser power distribution leads to microstructural and mechanical property variations that are still not well understood.

This work aims at better understanding the interaction of a 1 kW top-hat power distribution laser on a well know material, 316L stainless steel. The influence of texture and microstructure on relative density and crack density, when varying scan rotation, was evaluated. The high power (HP) laser and low power (LP) laser were compared with respect to microstructure and mechanical properties. HP leads to an increase in morphological and crystallographic texture together with a coarsening of the cell structure in contrast to the more random and finer cells found in LP processed material. Hot isostatic pressing was applied as a post-process treatment in order to close remaining pores and cracks. This helped in achieving higher elongations for LP and HP processed materials, while competitive mechanical properties to the 316L material specifications were obtained in both cases.

Keywords: High Power Selective Laser Melting; Additive manufacturing; Solidification microstructure; 316 L; Hot Isostatic pressing

1 Introduction

In the past years Selective Laser Melting (SLM), a powder bed based Additive Manufacturing (AM) process, is gaining interest in industrial sectors such as aerospace or energy. The flexibility in design as compared to traditional manufacturing and machining technologies and the capability of producing small batches have been the key drivers for this increasing interest. A highly focused laser source, with spot size between 40 and 120 μm [1]–[3], locally melts the metallic powder bed in a layer-by-layer fashion. However, this small laser spot size together with the concomitant penetration depth is the major determining factor of the most commonly used layer thickness which varies between 30 and 50 μm [4]–[6]. Hence the productivity is restricted. For this reason, a new generation of SLM processes is being introduced to the industry by the so called multi laser systems [2].

To increase the productivity, two main strategies are being followed: several 'standard' lasers working in parallel and the use of high power lasers (high power SLM, HP SLM). In the latter case, the productivity is increased thanks to the melting of thicker layers using the high power laser with up to 1 kW power.

However, not many studies have been performed on the material and laser interaction, and several challenges are expected to be encountered such as a reduced porosity and control or the existence of high thermal gradients leading to micro and macro cracking.

In 2011, Buchbinder et al. initiated the concept of HP SLM (high power SLM) for the production of AlSi10Mg parts using a 900 W laser obtaining densities of 99.5% while increasing the productivity by more than 300% [7]. Niendorf et al. showed that strong crystallographic textured columnar coarse-grains can be obtained by processing 316L SS with a top hat 1 kW laser, which led to an increased ductility compared to material processed by a 400 W laser [6]. Brenne et al. [8] analysed the microstructure of In718 with 1 kW laser after several heat treatments. In a recent work, Popovich et al. [9] analysed the tensile properties of In718 parts produced with regions scanned using either high or a low power laser. However, the effect of scanning strategies on parts that were manufactured with a high power laser has not been studied yet.

The use of higher power lasers may increase the thermal gradient and hence the cracking susceptibility. Moreover, the shape of the high power laser beam might differ from more common 'low power' lasers, hence influencing the heat input distribution and melt pool shape and dimensions.

Rapid solidification processes such as SLM make that materials are more prone to cracking. Typical reasons for crack formation are: the existence of very high thermal gradients (thermal

stress cracking), broadening of the solidification range during rapid solidification (solidification cracking or hot tearing) and the presence of rather brittle precipitates acting as stress concentrators (ductility dip cracking cracking). Especially certain aluminium alloys [10] and nickel based superalloys [4], [11]–[13] exhibit a high cracking sensitivity during SLM.

Although 316 L is considered as an easy-to-process material by SLM, cracking problems have been reported in previous works of laser metal deposition [14], [15] and also SLM [16], [17]. Several works reported on solutions for crack reduction or even their complete removal. Tomus et al. [18] and Harrison et al. [12] proved that by small compositional changes, particularly reducing the Si and Mn content, a significant reduction in crack density can be achieved for Hastelloy X. Montero et al. added silicon to Al7075 in order to avoid cracking thanks to a microstructural and solidification mode change [10]. Rickenbacher et al. [19] used hot isostatic pressing (HIP) in order to close the remaining cracks of IN738LC produced by SLM.

This work aims at better understanding the material and laser interaction for High Power SLM processes and its consequences on the part manufacturability, cracking and porosity control. To do so, the effect of microstructure and mechanical behaviour of high and low power SLM processed stainless steel 316L are investigated. Moreover, the influence of scanning rotations and HIP post-processing are analysed. The microstructure and crack density of different rotation angles between layers is analysed in as built and HIP condition. Light optical microscopy (LOM), scanning electron microscopy (SEM) and electron backscatter diffraction (EBSD) were used in order to study the grain structure and texture. Hardness, tensile and Charpy tests were performed in all samples to measure the mechanical performance of the different rotation angles and HIP treatment.

2 Material and methods

2.1 Starting powder

316L stainless steel powder provided by SLM Solutions Group AG (Lübeck, Germany) was used for this study, with powder particles sizes ranging from 10 to 45 μm and spherical in shape.

2.2 SLM process and scanning strategies

An SLM280^{HL} machine (SLM Solutions, Lübeck, Germany) equipped with a low power (400 W) and high power (1 kW) fibre laser was used to fabricate all SLM samples. The 400 W laser has a spot size of 76 μm with Gaussian distribution and the 1 kW laser has a 700 μm spot size with top hat power distribution.

Cube shaped samples with a volume of 1.2 cm^3 were built using a bi-directional scanning strategy. Three rotation angles between layers were used: 0°, 45° and 90° rotation (see Figure 1a). A core and shell scanning strategy was used for building the cubes, where the shell was scanned using low power (LP) laser (1 mm) and the core with the high power (HP) laser.

After SLM half of the samples were HIPed in order to reduce the cracks and porosity. The HIP process was conducted at Bodycote Hot Isostatic Pressing n.v. (Sint-Niklaas, Belgium) at 1000 bar and 1155 °C during 180 min. The different samples were labelled as follows: samples built with low power laser were named as LP-AB, and LP-HIP after the HIP process. Samples built with the high power laser were labelled HP-AB in the as built condition while after HIP, they were labelled HP-HIP.

The relative density was measured from the cross section images of polished surfaces using Image J software. An area of 1 cm² was analysed per sample in as built and HIP condition.

2.3 Phase identification and microscopy

All samples were sectioned parallel and perpendicular to the building direction (BD). After grinding, the samples were polished and further electrolytically etched using a solution of 10% oxalic acid at 7 V during 20 s. The microstructure was analysed using an Axiocam Leica light optical microscope (LOM) and a FEI XL-30FEG scanning electron microscope (SEM). The crystallographic orientation was investigated using electron backscattered diffraction (EBSD) which was performed on a FEI Nova NanoSEM 450 instrument equipped with a TSL/EDAX system. The step size used was 1.5 µm and the average confidence interval of the images was around 0.7. The relation between the crystallographic orientation and the colours in the EBSD images is shown in Figure 1b.

Figure 1. (a) Schematic representation of the applied angle of rotation in between layers: 0°, 45° and 90°. (b) Inverse pole figure which represents the relation between crystallographic orientation of the FCC austenite phase and colours in the EBSD images.

2.4 Mechanical properties

The mechanical behaviour of all samples was evaluated by tensile and Charpy V-notch testing. Rectangular tensile specimens were built in horizontal XZ direction and later machined to the final shape, according to the ASTM E 8M norm. The specimens were tested in an Instron 4505 device with a cross head velocity of 1 mm min⁻¹. For Charpy testing, rectangular bars were built in horizontal direction and machined to size and with V notch within XY plane. The test was performed according to ASTM E23.

Hardness testing was performed on a Future Tech FV-700 tester, applying a 0.5 kg load during 15 s. The XZ plane of all samples was investigated with five repetitions per sample.

3 Results

3.1 Microstructure comparison for LP and HP lasers

In order to compare samples built with low power (LP) and high power (HP), 90° rotation samples were etched to reveal the microstructure. They were analysed using LOM and SEM. The microstructure of LP-AB and HP-AB samples, analysed parallel to the building direction, is shown in Figure 2. For LP-AB, it can be seen that the melt pools are half cylindrical in shape. Grains grow across several layers exhibiting a width and length of 30-50 μm and around 100 μm , respectively (Figure 2a). For HP-AB, the melt pools are much wider and shallower. The columnar grains in the building direction can reach up to 1 mm in length and 100 μm in width. At the microscopic level, a difference in sub-grain structure size can be observed. The LP-AB sample presents smaller cell spacing than the HP-AB. In both conditions, cells growing perpendicular (Figure 2b,e) and parallel (Figure 2c,f) to the XZ plane are observed. Zhang et al. [20] also observed a gradual change from thin to coarse primary and secondary dendrite spacing with increasing the power in laser metal deposition of 316L. Saeidi et al. [21] observed a Mo enrichment at the sub-grain boundaries for ALM of 316L. For the HP laser, some cracks are visible for all samples in as built condition as also shown in Table 1. Cracks grow in between high angle grain boundaries as seen in Figure 2f.

Figure 2. Microstructural comparison between samples processed with standard low power LP-AB (a, b, c) and high power HP-AB (d, e, f) laser and with 90° between layers. (b, e) Show cells growing parallel and (c, f) perpendicular to the XZ plane. The yellow arrow indicate the cell growing direction.

3.2 Relative crack density

In order to reduce the cracking on HP-AB samples different cubes were built using varying scan rotations with the high power (HP) laser. Table 1 summarizes the relative density and crack density for parts LP-AB, LP-HIP, HP-AB and HP-HIP with respect to scan rotation. In the as built condition all HP-AB show densities below 99.6%. The 0° rotation results in the lowest relative density of 98.8%. After the HIP process, all densities increase reaching full densities close to 99.9%. In this case no significant spread is observed for all applied scanning strategies, contrary to the density values in the as built condition. For samples built with the low power, LP-AB, the densities obtained were 99.7% in the as-built state and increased to 99.8% for LP-HIP. The obtained density values were independent on the applied scanning strategy.

Table 1. Relative density of 316L in AB and HIP samples for 0°, 45° and 90° rotation processed by LP and HP.

	LP-AB		LP-HIP		HP-AB		HP-HIP	
	Rel. Density [%]		Rel. Density [%]		Rel. Density [%]	Crack density [mm/mm ²]	Rel. Density [%]	Crack density [mm/mm ²]
0°	99.7		99.8		98.8	0.58	99.9	-
45°	99.8		99.9		99.6	0.10	99.8	-
90°	99.7		99.9		99.6	0.09	99.8	-

After sectioning all samples parallel to the building direction the micrographs were analysed by LOM. The results are shown in Figure 3. For the HP-AB condition all samples show cracks parallel to the building direction. 0° rotation gives the highest crack density and longest cracks, with crack length varying between 23 and 1000 µm in size for 0°. All cracks are separated by 500 µm, which is the same distance as the hatch spacing used. For the other strategies, smaller crack lengths (between 30 and 300 µm for 45° and between 27 and 390 µm for 90°) are observed and no specific crack pattern is observed.

After HIP treatment, all cracks disappear for all samples, even for the long ones found in the samples processed using 0° rotation. Some remaining porosity is still present, but the pore fraction is significantly reduced compared to the AB condition.

Figure 3. LOM of high power laser in as built (HP-AB) (a, b, c) and Hot Isostatic Pressed (HP-HIP) (d, e, f) samples built with 0° (a, d), 45° (b, e) and 90° rotation angles (c, f) taken from the XZ plane.

3.3 Microstructure

The grain structure and local texture was analysed using electron backscatter diffraction (EBSD). Figure 4 shows the XZ plane for samples built with high power laser using 0°, 45° and 90° rotation angles.

Within these images an evolution on the grain morphology can be observed when different rotation angles are applied between two subsequent layers. For the 0° rotation columnar grains with 500 μm width in average can be observed, which equals the hatch spacing applied during SLM. The grain size is several millimetres in length. Some smaller grains can be observed in between the wide grains. When increasing the rotation angle, the wide grains break into thinner and slightly shorter grains. This is obtained due to a less pronounced solidification texture obtained during variation of scanning pattern from layer to layer. When increasing the scan rotation to 45° and 90°, the strong morphological texture decreases. The grains with the width of hatch spacing observed for the 0° scan rotation become 150 - 300 μm in width when subsequent layers are rotationally scanned. In addition, some grains appear to be tilted around 10 - 20° to the z axis.

Figure 4. EBSD images for samples built with HP laser and using 0° (a), 45° (b) and 90° (c) rotation angles from XZ plane.

Figure 5 shows the grain structure and local crystallographic orientation of HP-AB sample with 0° rotation from XY and XZ planes. On these images parallel tracks can be observed in both directions, parallel and perpendicular to the building direction. On the XY plane, grains with a width equal to the applied hatch spacing are observed (Figure 5a). Most of the grains grow normal to the scanning direction, along the horizontal thermal gradient, with a thickness around 100 µm, with rectangular beam shaped grains. The rectangular grains, seen from the XY plane, are decorated with equiaxed grains in the direction of the scanning track. A schematic representation of the XY plane is shown in Figure 5b representing two neighbouring tracks. Equiaxed grains are located at the centre of the melt pool and originate from the last fraction of liquid to solidify. This suggests that these grains were formed after the rectangular grains were created and quickly solidified. Epitaxial growth of the rectangular grains across the melt pools is observed. In addition, no macro-segregation was found between the edge (rectangular grains) and the centre-line (equiaxed grains) of the melt pool.

On the XZ plane of HP-AB with 0° rotation (Figure 5c) an evolution of the grain morphology and texture is observed between the high power laser and low power laser zone. It should be noted that all samples built with high power laser also have a contour of 1 mm around the whole sample. Therefore, the interface between LP and HP is shown at the bottom of Figure 5c. For the LP scanned area, small grains with a random orientation are observed at the bottom part. In the first 10 to 20 layers of the HP laser, no preferential orientation is observed. When the number of scanned layers is increased, competitive growth takes place evolving towards $\langle 100 \rangle$ crystallographic orientation in red colour which is the preferential solidification orientation for fcc austenitic stainless steels [22]. Grains are bigger than 1 mm in length and are confined to the

track width. Figure 5e shows a schematic representation of the XZ plane of two neighbouring tracks. It can be observed that cracks are formed in the centre of the scanned tracks, growing across several layers. The long and wide columnar grains are formed after two adjacent melt pools are scanned. The red arrows represent the different heat gradients for neighbouring tracks. When the interface between two melt pools is observed in a scanning electron microscope, different dendrite growth directions are observed (Figure 5d) within a grain with same crystallographic orientation. Dendrites are known to grow along the heat gradient [22] with 100 and besides having different growing direction, as seen in Figure 5d, epitaxial growth of the crystal is observed for them creating one single grain in between two neighbouring tracks. This again confirms that the equiaxed grains seen in Figure 5a are located at the centre of the melt pool, instead of at the melt pool borders.

Figure 5. Sample produced with HP laser in as built condition with 0° rotation. (a) Shows EBSD image of the XY plane and (b) the schematic representation of two neighbouring tracks depicting the equiaxed grains and rectangular grains. (c) EBSD images of the XZ plane and (d) SEM image of the zoomed region. (e) Schematic diagram of XZ plane of two neighbouring tracks showing the heat gradient direction with red arrows and the location of the cracks.

Figure 6 shows the vertical cross-section EBSD images of LP-AB, LP-HIP, HP-AB and HP-HIP samples with 90° rotation in between layers. The obtained grain size is significantly different between samples built with low and high power lasers, as previously observed in Figure 5c. Both conditions show grains that exhibit a morphological texture along the building direction. In addition, LP samples contain more equiaxed shaped grains, with an average width and length of 70 and 100 μm , respectively. In the case of HP samples, grains of more than a 1 mm in length can

be observed and less than 100 μm in width. After HIP, slight grain coarsening is observed for both conditions. However, the elongated grains are maintained for HP-HIP. LP-HIP shows less polygonal and more round shaped grains.

Local texture measurements obtained by EBSD are shown in Figure 6. LP-AB exhibits a weak $\langle 100 \rangle$ texture parallel to the laser scanning direction. After HIP, slightly more blue, hence slight evolution towards $\langle 111 \rangle$ direction texture is observed. Samples HP-AB and HP-HIP show a more enhanced preferential orientation of the $\langle 100 \rangle$ direction parallel to the scanning direction. HIP post processing does not seem to change this tendency. When comparing samples built with HP to LP, a stronger texture is observed for HP samples.

Figure 6 EBSD diffraction images of LP-AB (a), LP-HIP (b), HP-AB (c) and HP-HIP (d) taken from the XZ plane built with 90° rotation angle. The inverse pole figures are represented with respect to the XY plane and parallel to BD.

3.4 Mechanical properties' comparison for LP and HP lasers

Considering their higher relative density and reduced crack density values, samples built with 90° rotation angle were chosen to evaluate the mechanical properties for both LP and HP samples. The hardness evolution of all samples is shown in Figure 7. LP-AB samples show the highest hardness values with an average of 241 ± 6 HV which decreases to 212 ± 15 HV after applying HIP treatment. In high power samples the same tendency is observed, HP-AB samples exhibit a hardness of 213 ± 12 HV and HP-HIP 174 ± 6 HV.

Figure 7. Micro Vickers hardness of AB and HIP samples, processed with LP and HP.

Tensile tests were performed for all conditions, while horizontally built samples (thickness of the tensile test specimen in the Z direction) were used. The latter building direction is considered to be the most influenced direction by the presence of cracks. The tensile test results are shown in Figure 8a-c.

LP-AB shows the highest elongation values with an average of $47 \pm 4\%$ followed by LP-HIP with $39 \pm 10\%$ and HP-HIP with $37 \pm 10\%$. The lowest value was obtained for HP-AB with an elongation of $4.2 \pm 3\%$ at maximum ultimate tensile strength (UTS). Regarding the UTS, different tendencies are observed for low power and high power samples. LP-AB shows higher values than LP-HIP, while HP-AB is lower than HP-HIP. All samples except HP-AB show similar values and higher than the standard EN 10216-5:2013. In the case of yield tensile strength (YTS), as built samples (LP-AB and HP-AB) exhibit a higher value compared to HIPed specimens (LP-HIP and HP-HIP). In all cases the samples exceed the standard value of 190 MPa. The Charpy V-notch impact test results are shown in Figure 8d. For samples built with the low power laser, similar values are obtained; 126 ± 11 J for LP-AB and 120 ± 10 J for LP-HIP. HP-HIP exhibit only slightly lower impact energy values, averaging 111 ± 3 J and still higher than the standard EN 10216-5:2013. However, the impact energy value of the HP-AB sample falls below the standard.

Figure 8. Mechanical properties for samples built with LP and HP, in AB and HIP condition. (a) shows the elongation, (b) the ultimate tensile strength and (c) yield tensile strength obtained from the tensile testing. (d) shows the absorbed energy obtained from Charpy V-notch test. EN 10216-5:2013 was taken as reference standard.

4 Discussion

4.1 Effect of using high power laser

The results obtained in this work show the influence of the use of a high power laser with top hat energy distribution on the microstructure and mechanical properties of SLM processed 316L stainless steel. The use of this laser allows to increase the productivity rate 2.5 times thanks to the use of a higher powder layer thickness and hatch spacing, which is of high importance to widen the industrial acceptance of SLM as a near-net shape manufacturing technology.

Similar high densities are obtained for samples processed with a LP and HP laser. However, cracks occur when a high power laser is used. The increase in spot size, laser power and laser power distribution are the differences between both lasers. Their combination and their effect on

the heat input, and hence on the cooling trajectory, may have induced the cracking in HP and not in LP.

On the other hand, the resulting microstructure and properties are different compared to the ones obtained with lower energy lasers with a more concentrated and Gaussian shaped energy profile. Regarding the mechanical properties, HP-AB samples exhibit lower tensile elongation values, due to the presence of cracks that have propagated perpendicular to the tensile testing direction. In addition, hardness and yield tensile strength values are comparable to the ones obtained with the low power laser and much higher compared to conventionally produced 316L. Due to the early fracture of the HP-AB, ultimate tensile values are below the conventionally processed 316L stainless steel material. All properties can be significantly improved by HIP post processing, enabling the closure of remaining porosities and cracks.

4.2 Origin of the microstructure

The use of two geometrically different shaped laser beams with different power output profiles leads to the generation of differently shaped melt pools (Figure 2) which varies the heat gradient and hence the microstructure. When a low power laser with a focussed Gaussian beam profile is used to melt a 316L stainless steel powder bed, melt pools with narrow, deep and hemispherical in shape are created. Upon solidification, small columnar grains mixed with grains of random orientation are found. When a high power laser with a broad top-hat shaped energy profile is used, broad and shallow melt pools are formed. Upon solidification, large columnar grains, oriented parallel to the building direction, are observed.

At a microscopic level, no major differences are observed when the scanning rotation between subsequent layers is altered between 0 and 90°. The most significant microstructural differences were observed between the samples built with either the LP or the HP laser. Coarser dendrites were observed in HP laser processed samples as compared to LP laser processed samples. The increase in temperature gradient G decreases the primary arm spacing as proven in equation (1) [23].

$$\bar{\lambda}_1 = AG^{-a}V^{-b} \quad (1)$$

where $\bar{\lambda}_1$ is the average primary dendrite arm spacing, G is the temperature gradient (K/m) and V the solidification velocity (m/s). This suggests that the thermal gradient G increases when the LP laser is used, resulting in an increased cooling rate $G*V$ (K/s). Higher cooling rates induced by the LP laser reduce the cell spacing. Moreover, the grain boundary strengthening theory states that the grain size influences the yield strength since dislocation movement is hindered by the presence of a grain boundary. Since LP processed samples exhibit smaller grain sizes as well as smaller cellular sub-grain structures, higher strength and hardness values are observed in LP laser processed samples as compared to the HP laser processed samples, both in the as built and HIP

condition. Saeidi et al. [21] showed the high dislocation concentration at sub-grain boundaries of 316L produced by SLM, which further contributes to the higher hardness and strength values obtained for the LP laser processed samples.

Both the SEM and EBSD images, indicate that the 0° rotation scanning strategy, results in the generation of the most pronounced anisotropic microstructure among the different tested scanning rotation angles. After competitive growth has taken place, long columnar grains above 1 mm in length and 500 μm in width are found. This sample shows $\langle 100 \rangle$ preferential crystallographic orientation along the scanning direction. This microstructure resembles to the one observed in welding. Similar grain structures and grain orientations can be observed when processing at high welding speeds. In addition, comparable microstructures have also been observed for pure metallic Ta [5] and a βTi alloy [24] when produced by SLM applying a 0° rotation angle between subsequent powder layers. In welding, Arata et al. [25] obtained straight columnar grains normal to the welding direction and growing towards the centreline at 16.66 mm/s which is considered as a high welding speed. This morphology is obtained thanks to the teardrop shaped weld pool formed at high speeds compared to the elliptical weld pool obtained at low speeds [22]. Grains grow along the thermal gradient and perpendicular to the melt pool front. In a teardrop shaped weld pool, the trailing weld pool boundary is straight so that grains grow normal to the boundary and hence, point towards the centreline of the weld track [22], [26]. In the current work, the 700 μm wide, top hat shaped laser beam necessitates the application of lower scan speeds as compared to the LP laser processed samples. Nonetheless, independent on the applied beam shape, the cooling rates in SLM are still significantly higher when compared to welding.

Arata et al. [25] also found equiaxed grains around the centreline of the 1050 aluminium alloy produced at high weld speeds, called axial grains. In Figure 5a also small equiaxed grains can be seen in the middle of the melt pool for the AB 0° . Those grains might have formed at the tail of the melt pool, at the cooler melt pool surface.

The columnar grain morphology is not obtained for other rotating angles even when the same scanning parameters were applied within a single layer. This suggests that the repetitive and constant heat pattern across the layers is favourable for the formation of the columnar straight grains growing normal to the scanning direction. When rotating the scanned vectors over the layers, the orientation of the induced thermal gradient is more variable, hence continuously altering the preferred solidification orientation and resulting in the formation of smaller grains. This smaller grains are favourable in the reduction of cracking, as confirmed by the crack density calculated in Table 1.

4.3 Effect of HIP

Hot Isostatic Pressing is commonly used as a post processing technique for SLM parts in order to reduce the porosity as a heat treatment among others[19]. In the current work, HIP has proven to be a good choice to eliminate the remaining porosities and cracks formed during SLM processing. The typical sub-grain dendritic structure observed in SLM processed 316L stainless steel, disappears after HIP post-processing. Slightly coarser grains are observed for both LP and HP lasers after HIP treatment. However, the elongated grains are maintained for HP-HIP and LP-HIP exhibiting less polygonal and more rounded grain morphologies.

Elongation at break reduces slightly after HIP for samples built with LP laser. After HIP, samples were cooled down slowly which could have caused sigma phase formation around 800 °C, which can decrease the elongation of 316L due to its brittle nature [27]. However, its presence could not be detected neither by EDS nor by X-ray diffraction. On the contrary, elongation increases abruptly for HP-HIP to values close to LP-HIP thanks to the crack removal of LP-AB. YTS and hardness decrease compared to the as built condition which is related to the complete dissolution of sub-grain cellular structure. Sub-grain microstructure differences between LP-AB and LP-HIP are shown in Figure 9. This effect was also observed in a previous work where heat treatments at temperatures above 950°C removed the cellular dendritic structure hence lowering the hardness values [28]. Nevertheless, hardness and yield values still exceed the ones from conventionally produced 316L. UTS decreases for LP-HIP while it increases for HP-HIP. This can be related to the early fracture of HP-AB due to the presence of cracks, resulting in lower UTS values than expected.

Therefore, HIP treatment proves to be a fruitful solution for removing cracks while maintaining the mechanical properties for components that need faster production times using a 1 kW laser.

Figure 9. SEM micrograph of LP-AB (a) and LP-HIP (b) samples showing sub-grain features.

5 Conclusions

The use of high power selective laser melting (HP-SLM) is shown to be a good choice for increasing 2.5 times the standard SLM process productivity rate.

The use of a high power (HP) laser (1 kW, beam radius 700 μm) with a top hat power distribution generates a wide and shallow melt pool shape, as compared to a low power (LP) laser (400 W, beam radius of 70 μm) with a narrow Gaussian power distribution, hence changing the size and orientation of the thermal gradient within the melt pool. The existence of wider and shallower melt pools results in the formation of long columnar grains growing along the building direction, when no scan rotation is used in between subsequent powder layers.

Rotating the scanning angle between powder layers changes both the grain morphology and the grain crystallographic orientation for HP laser processed samples. When no scan rotation is applied in between subsequently deposited powder layers, a strong $\langle 100 \rangle$ texture along the scanning direction is observed. Simultaneously, however, the crack density is highest amongst all explored scan settings, with cracks located in the centre of the melt pools and growing parallel to the build direction. Therefore, a rotation angle of 90° was chosen for mechanical property investigation and optimisation.

When a low power (LP) laser with a narrow Gaussian power distribution is used, smaller and more randomly oriented columnar grains are formed. The use of a low power laser induces larger thermal gradients leading to a higher cooling rates and hence a finer sub-grain microstructure. This leads to an increase in Vickers hardness as well as in UTS and YTS values.

Hot isostatic pressing enables the removal of the remaining pores and cracks in all laser manufactured samples. Mechanical properties such as elongation and ultimate tensile strength improve significantly for HP-HIP, as all cracks are completely closed. On the other hand, yield strength, hardness and ultimate tensile strength decrease slightly after HIP. Nevertheless, all mechanical properties for both lasers exceed the ones from the specifications when applying a HIP post processing treatment.

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