Strain Localization in Ti-6Al-4V Widmanstätten Microstructures Produced by Additive Manufacturing

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Abstract

Although Additive Manufacturing (AM) offers numerous performance advantages over existing methods, AM structures are not being utilized for critical aerospace and mechanical applications due to uncertainties in their structural integrity as a result of microstructural variations and defects arising from the AM process itself. With strain localization serving as a precursor for material failure, the impact of microstructural variations in Direct Metal Laser Sintering (DMLS) produced Ti-6Al-4V on strain localization is investigated through high resolution digital image correlation as a result of monotonic and cyclic loading for both stress relieved and mill annealed specimens. From the experimental characterizations, it was observed that the former β phase boundaries that enclose separate microtextured regions and the α lath's major axis orientations play a critical role in the localization of strain with respect to the Widmanstätten microstructures exhibited by DMLS Ti-6Al-4V. The results show a potential opportunity to improve the performance of DMLS produced Ti-6Al-4V by decreasing the size of the β grains during the solidification process through higher cooling rates and a build strategy that avoids the reheating of solidified material as much as possible to prevent β grain growth.

Keywords: microstructure; digital image correlation; strain localization; Ti-6Al-4V; additive manufacturing; prior β boundaries.

Introduction

With Ti-6Al-4V being the most widely used titanium alloy [1], there is a significant interest to access the uncertainty associated with the material properties of Direct Metal Laser Sintering (DMLS) produced Ti-6Al-4V. This uncertainty is almost exclusively driven by the variability of the DMLS process itself. With strain localization acting as a precursor to crack initiation and ultimately failure [2,3], an understanding of the microstructural impact on the localization of strain is necessary to understand its role in the failure of DMLS produced Ti-6Al-4V under monotonic and cyclic loading conditions.

Microstructures for additive manufacturing (AM) built materials are typically fine, non-equilibrium phase structures due to the rapid solidification and steep thermal gradients inherent to the AM process. For AM produced Ti-6Al-4V, a martensitic, α' , microstructure is initially

observed due to steep thermal gradients and rapid cooling from above the martensite start temperature [4]. To remove the AM martensitic microstructure in Ti-6Al-4V, a post-processing heat treatment, below the β transus temperature, in the $\alpha + \beta$ region of the Ti-6Al-4V phase diagram is utilized to refine the microstructure. When annealed, the martensitic, α' microstructure is transformed to an $\alpha + \beta$ microstructure with the precipitation of incoherent β particles among the α lathes or as β layers around α lathes [1].

Most current research focuses on the fatigue performance of additively produced Ti-6Al-4V at the macroscale. In selectively laser manufactured Ti-6Al-4V, it is postulated that crack initiation is linked to internal porosity defects in the material as a result of the AM process, and that crack growth behavior is linked to internal residual stresses in the material that can get quite high as a result of the AM process [5]. To improve performance in AM Ti-6Al-4V, post-processing normally includes a thermal stress-relief treatment to improve crack growth behavior and a hot-isostatic pressing treatment to minimize porosity and improve crack initiation [5].

Due to the wide range of industrial applications, Ti-6Al-4V exists in various microstructural forms, which are engineered to achieve desired performance. Industrially significant Ti-6Al-4V displays a complex two-phase microstructure, $\alpha + \beta$, over a wide range of length scales, which exhibit intricate deformation mechanisms [6]. At room temperature, the prismatic slip system in the α phase displays the lowest critical resolved shear stress [7]. At increased concentration of Al solute atoms and temperature, basal slip becomes more prevalent, with basal and prismatic slip being equally likely at 5.74% (by weight) [8].

In the 1970s, researchers have discussed the formation of fatigue cracks in Ti alloys along slip planes that exhibit intense shear [9,10], which has been observed to be either basal or prismatic planes [11], although more commonly cracks first initiate on the basal planes [12]. Stubbington and Bowen proposed that the fatigue life of Ti-6Al-4V can be improved by reducing the size of the α colonies, i.e. the α lath length (size of the colonies), leading to smaller nucleation crack sizes [10]. Echlin et al. observed via strain mapping of equiaxed Ti-6Al-4V that basal slip was observed first, followed by slip along prismatic and pyramidal systems with increasing strain [13].

For α Ti alloys, grain combinations of hard-soft α grains, i.e. grains that are favorably oriented for slip adjacent to grains with low slip tendencies, are prone to crack initiation, especially in dwell fatigue conditions, as proposed by Evans and Bach [14] and simulated by Ghosh and co-workers [15] and Dunne and Rugg [16,17]. Through in-situ studies of Ti-7Al, Brandes pointed to grain-to-grain variation in slip character as a driving force for crack initiation [18], which was extended by Chatterjee et al. to the role of triaxiality as measured by high energy x-ray diffraction microscopy [19]. For Ti-6Al-4V, the role of the β boundaries influence failure, as studied by Pilchak et al.; their results indicate that boundaries which are crystallographically suited for slip transfer tend to initiate fatigue cracks [20].

In this study, strain localization of Ti-6Al-4V produced by DMLS¹ is explored by means of high-resolution digital image correlation (HR-DIC) as a precursor to crack initiation. The use of HR-DIC allows for more complete information on slip activation relative to the microstructure, as compared to slip trace analysis [13]. In this study, conditions of Ti-6Al-4V in the as built (with a stress relief process) and mill annealed condition, both displaying a

¹ DMLS is a trade name associated with this process, while selective laser melting is a more accurate description of the process used in this study.

Widmanstätten microstructure, are examined after monotonic and cyclic loading. The role of microtextured regions and the prior β boundaries are discussed, pertaining to strain localization.

Materials and Methods

The Ti-6Al-4V test specimens for this research were manufactured on an EOSINT M280 system at Penn State University's Center for Innovative Materials Processing through Direct Digital Deposition (CIMP-3D). The test specimens are 3 mm thick and 50.8 mm in overall length by 8.9 mm in width, while the gauge section is 17.5 mm long by 3.5 mm in width. Half the specimens were built in a horizontal direction and the other half of the specimens were built in a vertical direction (Fig. 1). Monotonic loading specimens were built net shape, while cyclically loaded specimens were EDM cut from bulk blocks. Two post processing thermal treatments were applied to different specimens. The first treatment referred to as a stress relief treatment was performed by Bodycote in a vacuum/nitrogen atmosphere at 650 °C for three hours. The second treatment referred to as post-processed was a mill anneal performed by Solar Atmospheres in a vacuum/argon atmosphere at 704 °C for two hours followed by argon cooling. It is noted that both the stress relief and mill anneal heat treatments are both within the $\alpha + \beta$ region of the Ti-6Al-4V phase diagram and in this regard, the heat treatments are rather similar in nature.

Upon receipt, specimens were ground using 320, 400, 600, 800, and 1200 grit paper followed by polishing on both rotary and vibratory polishers with a solution consisting of four parts 0.05 μm colloidal silica and one part 30 % solution of hydrogen peroxide. An area of interest for each specimen was marked using fiducial markers (in the form of Vickers microhardness indents at 25 g force) in a 70 μm square pattern with one side having an additional marker for orientation purposes. The fiducial markers are required to consolidate microstructure and strain data. After cleaning with acetone and methanol, the area marked with fiducial markers was characterized via electron backscatter diffraction (EBSD) using an FEI XL-40 scanning electron microscope (SEM) with an EDAX EBSD camera and associated data collection and analysis software. The SEM setup parameters included a 25 kV accelerating voltage, a spot size of 5, a 100 μm aperture, a working distance of 17 mm, and a 70° tilt were utilized. The EDAX EBSD setup parameters included a gain of 12.60, a black of 1.51, a 31.97 ms exposure time, and a step size of 0.20 μm . Due to limitations with using a smaller step size with the current SEM/EBSD setup and the small size of the β phase, only the α phase was probed during EBSD using a material file for hexagonal symmetry with lattice parameters of $a = 2.95 \text{ \AA}$ and $c = 4.68 \text{ \AA}$. Backscatter electron (BSE) imaging for cyclically loaded specimens was performed on a FEI Quanta 3D scanning electron microscope at 10,000x magnification with an accelerating voltage of 15 kV, a spot size of 4.5, a working distance of 5 mm.

High resolution digital image correlation (HR-DIC) was performed on each of the specimens. The samples were speckled with a self-assembled Au nano-particles as described by Kammers and Daly [21] with procedure outlined in [22]. Soak time for the Ti-6Al-4V specimens in the Au nano-particles solution varied from 5 to 10 days to produce a dense enough correlation pattern. For the high resolution digital imaging correlation, a FEI Quanta 3D scanning electron microscope at 8,000x magnification was utilized with an accelerating voltage of 5 kV, an aperture size of 30 μm , a spot size of 5 and a working distance of 10 mm. Conducting HR-DIC within an SEM is subject to distortion of the captured images based on the electromagnetic fields

present within the SEM [23,24]. Details of the SEM parameter optimization, spatial distortion correction, drift minimization, and ex-situ procedure can be found in [25]. VIC2D [26] was utilized to correlate Ti-6Al-4V strain maps from the SEM images with a step size of 0.037 μm and a subset size of 1.381 μm for as-built monotonic specimens and 1.749 μm for bulk cyclically loaded specimens.

Monotonic Loading Results

The as-built Ti-6Al-4V specimens were subjected to monotonic loading to a strain of 2.5% (as measured by an extensometer) and unloaded to zero load, as shown in Fig. 2. The yield and hardening response of all four specimens are similar, with the recorded residual strain upon unloading measured at 1.56%, 1.50%, 1.54%, and 1.52% for the four test conditions: (i) horizontal build / stress-relieved, (ii) horizontal build / post-processed (in the mill annealed condition), (iii) vertical build / stress relieved, and (iv) vertical build / post-processed (in the mill annealed condition), respectively. For all microstructure characterization, the data was partitioned to discard probe points with confidence intervals less than 0.10 and the EBSD inverse pole figure (IPF) maps for these samples are shown in the first column of Fig. 3.

All four samples display an α phase Widmanstätten microstructure [27]. During sintering and subsequently solidification, the material is above the β transus temperature and the cooling rate is sufficient for the individual α -phase lamellae to grow into the prior β -phase grains, nearly instantaneously. That is, a phase transformation exists during cooling, where the cubic β grains transform into 12 distinct crystallographic variants of the hexagonal α phase. As shown in Fig. 3, the strain component corresponding to the axial loading, ϵ_{xx} , is depicted based on the DIC analysis. It is noted that the axial strain component is dominant, as expected, and can be quite simply used for comparisons to potential modeling efforts. For this reason, ϵ_{xx} values are used to plot the strain maps throughout this paper. Examining the strain map (for the first, third, and fourth cases corresponding to the rows in Fig. 3), almost all strain localizes itself in long α lathes at 45° angles from the horizontal tensile direction in accordance with the highest shear strain at the macro scale. In the second sample (horizontal build/post processed condition), strain localization occurs in long α lathes, whose major axis is at a 90° angle to the horizontal tensile direction. All monotonically loaded strain maps are scaled from 0 to 5% in the loading direction (horizontal direction). For the monotonic strain condition, some out of plane grain movement was observed after loading and confirmed using a Zygo ZeGage Optical Surface Profiler, so the strain maps presented for the monotonically loaded cases represent a 2D projection of 3D strain.

All microstructure and strain maps have the suspected former β phase boundaries marked by white dash line, as shown in Fig. 4 (DIC images with former β phase boundaries shown in the first column). Former β phase boundaries are identified by a combination of changes in texture and grain orientation [27]. These regions inside the former β phase boundaries can be referred to as microtextured regions (MTR), as each region is observed to have its own texture and orientation preference [28]. A critical resolved shear stress (CRSS) ratio of 1.05 between the basal slip systems $\{0001\} <12\bar{1}0>$ and the prismatic slip systems $\{10\bar{1}0\} <12\bar{1}0>$ is utilized [29]. The Schmid factor maps for the prismatic slip systems are shown in the second column of Fig. 4 and the basal Schmid factor maps are shown in the third column.

The former β phase boundaries appear to have some impact on the strain localization as some strain discontinuities are observed at these boundaries, while at other places on the

boundaries strain appears to transmit through the β boundary. Observing the Schmid maps in Fig. 4, it appears that all strain in the α lathes is coincident to high Schmid factors observed in the prism slip systems rather than basal slip system. For the fourth monotonically loaded specimen (vertical build / post-processed condition), vertically oriented strain localizations are observed within the microstructure contained by the former β phase boundary in the upper right corner of Fig. 4(j). However, in Fig. 4(l), it is observed that 45° oriented α lathes accumulate strains according to a high Schmid factor for basal slip, while those vertically oriented α lathes exhibit strains contained by the upper right former β phase boundary coincide with high Schmid factor for prism slip. It is also observed that the microstructure contained by the upper right former β phase boundary appears to have a higher overall Schmid factor for prism slip than the rest of the microstructure, which indicates that the basal slip may be a result of grain interactions to accommodate the compatibility of deformation.

Fatigue Loading Results

The dogbone Ti-6Al-4V specimens were EDM cut from bulk blocks and ground. Specimens were first loaded to 1% strain (measured by an extensometer) at a rate of 10^{-4} to initialize some plasticity in the specimens. Recording the maximum stress at 1% strain, the specimens were then cycled using that maximum stress and a stress ratio of $R_\sigma = 0.05$ at 1 hz for 1000 cycles. Upon unloading to zero force after 1000 cycles, the residual strain was recorded as 0.43%, 0.49%, 0.45%, and 0.47% for the cyclically loaded test conditions: (i) horizontal build / stress-relieved, (ii) horizontal build / post-processed (in the mill annealed condition), (iii) vertical build / stress relieved, and (iv) vertical build / post-processed (in the mill annealed condition), respectively. To get an estimate of fatigue life, specimens from two test conditions were cycled using the above fatigue profile to failure. Fatigue lives for the two vertical build conditions were found to be pretty similar, as the vertical build/stress relief condition failed at 3,611 cycles and the vertical build/post-processed condition failed at 3,400 cycles. This means that a thousand cycles represents about a quarter to a third of the specimen's overall fatigue life. Conducting a fracture analysis within the SEM (please refer to the Appendix), it was observed that both specimens failed from surface flaws that propagated through the cross section despite the specimens undergoing grinding using 600 grit paper to remove any pre-existing surface roughness. For all microstructures characterized, the data was partitioned to discard probe points with confidence intervals less than 0.10, as the EBSD IPF plots are shown in the first column of Fig. 5. The second column of Fig. 5 displays the BSE images, as the contrast is the best way to indicate the β phases (depicted as white).

All four samples display an α phase Widmanstätten microstructure [27]. All cyclically loaded strain maps are scaled from 0 to 2.5%. For all four samples, strain localization is primarily observed at a 45° angle from the horizontal loading direction, which coincides at the macro scale with the maximum shear strain. Additionally for the two stress relieved samples, Fig. 5(c,i), strain localizations also appears vertically at 90° to the horizontal loading direction. For the cyclic loading condition, no out of plane movement was observed after the completion of cyclic loading using a Zygo ZeGage Optical SurfaceProfiler, except for Fig. 5(f), which corresponds to the horizontal build / post-processed sample (in the mill annealed condition), so the strain map for this test condition represents a 2D projection of 3D strain. It is noted that the location of the strain maps are in the middle of the specimens, thus ensuring that the strain maps are similar in nature between the monotonic (manufactured near-net shape) and fatigue

(machined from bulk blocks) samples, in which the microstructure is sufficiently far away from the free surface of the part during the AM build process.

All microstructure and strain maps have the locations of suspected former β phase boundaries marked by white dash lines to highlight microtextured regions (first column of Fig. 6). The Schmid factor maps for the prismatic slip systems are shown in the second column of Fig. 6 and the basal Schmid factor maps are shown in the basal slip system, with the ratios of CRSS fixed as previously mentioned.

For the first sample, horizontal build / stress-relieved condition, at the left former β grain (Fig. 6(a)), strain is not observed to cross the boundary, while it does cross the boundary at the center of the image. In the second specimen, horizontal build / post-processed (Fig. 6(d)), it is observed that strain localization is influenced by the left former β phase boundary, meanwhile the strain path transmits past the former β phase boundary on the right. In the third sample, vertical build / stress relieved conditions (Fig. 6(g)), strain is not observed to cross former β phase boundaries when slip impinges at the boundaries. Finally, in the fourth sample, corresponding to the vertical build / post-processed condition in Fig. 6(j), the strain path is impeded at the right former β phase boundary but is nearly continuous across the left boundary.

In each of these conditions, it is observed that in cases where the strain path is impeded by the former β phase boundary, the major axis of the α lathes on each side of the boundary display a large misorientation. Meanwhile for cases where the strain path is transmitted through the former β phase boundary, the major axis of the α lathes are more closely aligned to each other. In all four samples, slip localization corresponds to regions with high Schmid factors on the prismatic as well as basal slip systems.

Discussion

As shown in Figs. 4 and 6, strain maps are measured by HR-DIC. With the advent of orientation imaging microscopy (OIM), it is straightforward to compare the strains with the Schmid factor maps. The Schmid factor maps project the applied macroscopic stress onto the slip systems rotated for the orientation of each grain. This implies that the stress on each grain is the same as the macroscopic stress, known as an isostress condition according to the Sach's model [30]. In reality, there are elastic and plastic anisotropies resulting from grain interactions within the polycrystalline aggregate. The anisotropic elastic stresses and strain incompatibilities between grains results in complex stress states in each grain. Bernier et al. highlighted the importance of using the stress tensor in each grain to determine the active slip systems and not the macroscopic applied stress [31]. In agreement with this statement, it has been shown that there is not a correlation between the Schmid factor and high strain values measured by DIC in FCC materials [32]. In the present study, it is interesting that a high correlation is observed between the Schmid factor maps of the prismatic and basal slip systems of the α phase and the HR-DIC strain localization in $\alpha+\beta$ Ti-6Al-4V. Similarly, in $\alpha+\beta$ Ti-6Al-4V, Bridier et al. suggested that the Schmid law was satisfactory to explain slip system activation with a combination of OIM and slip trace analysis [33]. One possibility for this agreement is the role of the β phase in between the α lathes in reducing grain interactions between the α lathes. In the monotonic loaded specimens, the regions of high strain localization exhibited high maximum Schmid factor for the prismatic slip systems within the α lathes. For the cyclic loaded specimens, both high Schmid factors on the prismatic and basal slip systems corresponded with

regions of high strain accumulation, as basal slip has been observed to occur before macroscopic yielding [13] and is seen to be more prevalent in the cyclic loading cases.

Further, the results show cases of slip transmission across the microtextured regions (MTRs), i.e. former β phase boundaries. In the monotonically loaded cases, more slip transmission was observed. While in the cyclically loaded case, the specimen was subjected to less macroscopic accumulated strain, and as such, there are more cases of slip being impeded by the former β phase boundaries (across the MTRs). Lee, Robertson, and Birnbaum's (LRB) strain incompatibility criteria for slip transmission across a grain boundary (GB) were developed for FCC materials [34]. The first is a geometrical condition for predicting the active slip plane by minimizing the angle between the slip directions and line of intersection of the slip planes to the GB. The second part of the criterion determines the slip direction based on the maximum resolved shear stress on the active slip plane. The third condition requires a Burgers vector, $\mathbf{b}$, of the residual dislocation (difference in $\mathbf{b}$ of incoming and outgoing dislocation) to be minimized. Kacher and Robertson observed these criteria hold for α -Ti, with the residual Burgers vector governing the dislocation mechanics of slip transmission [35]. In OIM, it is not always practical to employ the LRB criteria, due to lack of information of the GB inclination plane. Bieler et al. used a geometric criterion, m' , to identify slip transmission in Ti alloys. They showed concurrence between $m' = \cos \kappa \cdot \cos \psi$ and the slip transfer events, where κ is the angle between slip vectors and ψ is the angle between slip plane normals [36.]. In the present work, slip transmission across the MTRs (former β phase boundaries) are observed in cases where the orientations of the α lathes are aligned, which is in agreement with the m' criterion for alignment of the slip systems. While within the MTRs, slip occurs along the major axes of long α lathes, oriented approximately 45° to the macroscopic loading axis, i.e. having a maximum resolved shear stress. It has been shown for Widmanstätten microstructures that the α lathes may share common basal planes relative to a $\sim 10.5^\circ$ rotation about the $\mathbf{c}$ -axis of the crystal [37]. This crystallographic relationship is interesting to note, since within the MTRs, many of the α lathes share common characteristics of slip system activation.

Lastly, the four cases, based on build direction and heat treatment, produce similar strain localization features. This would imply that the build direction and heat treatment has limited influence on the microstructure and slip characteristics. In this study, a small region was analyzed for each case, $70\ \mu\text{m} \times 70\ \mu\text{m}$, as an emphasis was placed on high-resolution images and as a consequence the region of interest was small. Similar strain distributions are expected across the sample; albeit a statistical analysis across various MTRs is needed for definitive conclusions of the role of build direction and heat treatments, including analysis of the potential role of porosity.

Conclusion

The following conclusions can be reached in regards to strain localization for Ti-6Al-4V Widmanstätten microstructures created through the DMLS build process and subsequent thermal refinement.

1. Former β phase boundaries play a critical role in strain localization within the microstructure of DMLS produced Ti-6Al-4V parts. Each of these former β phase boundaries represents a separate microtextured region with its own texture and grain major axis orientation preference. If several microtextured regions are oriented to allow

slip system activation and subsequently slip transmission across the former β phase boundaries, then large accumulations of strain will be observed. However, if strain is impeded at each former β phase boundary preventing the transmission of slip, then the accumulation of larger strains necessary for crack initiation can be delayed.

2. It was observed that strain manifests along the major axis of α lathes as the major axis within an α lath provides a route relatively free of barriers for slip, such as grain or phase boundaries. Strain accumulations are observed to occur in large α phase lathes with favorability oriented slip systems that have a major axis closely aligned at approximately a 45° angle to the loading direction, which corresponds to the highest macro scale shear strain.
3. Slip is seen to transmit across microtextured regions (MTRs) with alignment of the α lathes on both sides of the former β phase boundaries, based on a geometric argument of slip systems across the adjacent grains. As the macroscopic strain is increased, more slip transmission events are observed across MTRs.
4. The regions of high strain accumulation as measured by HR-DIC are in agreement with the Schmid factor maps for the prismatic slip systems in the α lathes. For the cyclic loaded specimens, strain accumulation on the basal slip system in the α lathes is more prevalent. The Schmid factor maps do not account for the complex stress states in each grain due to grain interactions and compatibility within the polycrystal, so it is rather surprising that the strain accumulation follows region of high Schmid values.
5. Opportunity exists to improve the performance of DMLS produced Ti-6Al-4V by controlling the size of the former β grains early in the solidification process. By decreasing the size of the β grains during the solidification process by increasing the cooling rate and a build strategy that avoids the reheating of solidified material as much as possible to prevent β grain growth, the size of the microtextured region can be decreased to resist slip between microtextured regions (similar to grain boundary strengthening).

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Appendix

As earlier discussed, the monotonic samples were produced in a net shape condition via DMLS, while the cyclic samples were EDM cut from bulk blocks produced via DMLS. The surface roughness for each sample was measured using a Zygo ZeGage optical surface profiler and the results are shown in Table 1. As expected, the DMLS samples produced in the net shape

condition were significantly rougher than the EDM surface finishes, especially in the vertical build direction. In order to avoid premature failure of the cyclic samples from surface imperfections, the cyclic samples were chosen from the interior of bulk blocks (these results are expressed in Table 1). Afterwards, the cyclic samples were ground using 600 grit paper to further reduce the surface roughness. Fracture surface of two fatigue samples are shown in Fig. 7. As can be seen in Fig. 7, multiple origins of crack initiation are apparent, with the black arrows pointing to regions that experienced significant crack growth. All cracks grew from the imperfections that were at the surface or near the surface. Also, the fracture surfaces are similar in nature between the samples that were stress relieved and post-processed, except more crack initiation sites are apparent in the sample that only underwent a stress relief heat treatment.

Table 1. Surface roughness for the monotonic samples produced via DMLS in the near-net shape condition and cyclic samples that were EDM cut from blocks produced via DMLS.

	Surface Roughness (μm)	
	Near-net shape (monotonic) samples	Machined from bulk (cyclic) samples
Horizontal Build / Stress Relieved	4.61 ± 0.90	3.85 ± 0.16
Horizontal Build / Post-Processed	4.15 ± 0.62	3.90 ± 0.21
Vertical Build / Stress Relieved	20.45 ± 2.83	3.85 ± 0.15
Vertical Build / Post-Processed	11.70 ± 1.24	3.72 ± 0.13

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HR-DIC Strain Maps for DMLS Ti-6Al-4V

Slip transmission across prior β boundaries

Slip localization at prior β boundaries

Figure 1. Build direction of the dogbone specimens displaying the area of interest.

Figure 2. Monotonic stress-strain curves for the Ti-6Al-4V samples in various conditions. The samples are loaded to a maximum strain of 2.5% and then unloaded to zero force.

Figure 3. Characterization of the net-shaped dog bones that were subjected to monotonic loading. The first, second, third, and fourth rows correspond to (i) horizontal build / stress-relieved, (ii) horizontal build / post-processed (in the mill annealed condition), (iii) vertical build / stress relieved, and (iv) vertical build / post-processed (in the mill annealed condition), respectively. The first column is an IPF map of the α phase from EBSD and the second column represents DIC strain maps with respect to the horizontal loading direction (strain scale bar shown at the bottom).

Figure 4. Analysis of the net-shaped dog bones that were subjected to monotonic loading, with prior β boundaries shown with a white dash outline. The first, second, third, and fourth rows correspond to (i) horizontal build / stress-relieved, (ii) horizontal build / post-processed (in the mill annealed condition), (iii) vertical build / stress relieved, and (iv) vertical build / post-processed (in the mill annealed condition), respectively. The first column represents DIC strain maps (strain scale bar shown at the bottom), while the second and third columns are the Schmid factor maps for the prismatic and basal slip systems, respectively.

Figure 5. Characterization of the fatigue specimens after 1000 cycles. The first, second, third, and fourth rows correspond to (i) horizontal build / stress-relieved, (ii) horizontal build / post-processed (in the mill annealed condition), (iii) vertical build / stress relieved, and (iv) vertical build / post-processed (in the mill annealed condition), respectively. The first column is an IPF map of the α phase from EBSD, the second column is a BSE image displaying the α / β phases, and the third column represents DIC strain maps with respect to the horizontal loading direction (strain scale bar shown at the bottom).

Figure 6. Analysis of the fatigue specimens after 1000 cycles, with prior β boundaries shown with a white dash outline. The first, second, third, and fourth rows correspond to (i) horizontal build / stress-relieved, (ii) horizontal build / post-processed (in the mill annealed condition), (iii) vertical build / stress relieved, and (iv) vertical build / post-processed (in the mill annealed condition), respectively. The first column represents DIC strain maps (strain scale bar shown at the bottom), while the second and third columns are the Schmid factor maps for the prismatic and basal slip systems, respectively.

Figure 7. Fracture surface analyses of the fatigue samples: (a) vertical build / stress relieved and (b) vertical build / post-processed (in the mill annealed condition). A bulk block is produced via DMLS and the test specimens are machined from the block.

Figure 1.

Figure 2.

Figure 3.

Figure 4.

Figure 5.

Figure 6.

Figure 7.