Fatigue crack growth mechanisms at the microstructure scale in as-fabricated and heat treated Ti-6Al-4V ELI manufactured by electron beam melting (EBM)

Haize Galarraga¹, Robert J. Warren^{*1}, Diana A. Lados¹, Ryan R. Dehoff^{2,3}, and Michael M. Kirka^{2,3}

¹ Worcester Polytechnic Institute, Integrative Materials Design Center, 100 Institute Road, Worcester, MA 01609, USA

² Materials Science and Technology Division, Oak Ridge National Laboratory, Oak Ridge, TN 37831, USA

³ Manufacturing Demonstration Facility, Oak Ridge National Laboratory, Knoxville, TN 37932, USA

*Corresponding Author: E-mail address: rwarren@wpi.edu

Abstract

Electron beam melting (EBM) is a metal powder bed fusion additive manufacturing (AM) technology that fabricates parts by selectively scanning consecutive powder layers with an electron beam. Additive manufacturing technologies are increasing in importance for aerospace and medical applications, where the demand for a fundamental understanding and predictability of static and dynamic material properties are high. Ti-6Al-4V is the most widely used and studied alloy for this technology, and is the focus of this work in its Extra Low Interstitial (ELI) variation. The layered manufacturing of metallic components by EBM creates a unique directional microstructure, and consequently, anisotropic properties. Microstructure evolution, and its influence on mechanical properties of the alloy in the as-fabricated condition, has been documented by various researchers. However, fatigue crack propagation and the effects of the directional structure have not been sufficiently studied, imposing a barrier for this technology's potential extension to high-integrity applications. In this study, fatigue crack growth (FCG) both parallel and perpendicular to the build directions was studied for different stress ratios and crack growth stages. The interaction between the directional as-fabricated EBM microstructure and FCG was investigated and compared to that of the equiaxed β annealed microstructure obtained by annealing above the β transus temperature. The FCG threshold, ΔK_{th} , was analytically modeled for the two relative crack propagation directions at different stress ratios, and FCG microstructural mechanisms were established for all three regions of crack propagation.

KEYWORDS: additive manufacturing; electron beam melting; Ti-6Al-4V; fatigue crack growth mechanisms; threshold modeling

1 Introduction

Electron beam melting (EBM) is a metal powder bed fusion additive manufacturing (AM) technology used for the fabrication of three-dimensional near-net-shaped functional components directly from computer models [1]. The ASTM F2792 standard defines AM as “a process of joining materials to make objects from 3D model data, usually layer upon layer, as opposed to subtractive manufacturing methodologies” [2]. The variety and quantity of applications for AM technologies, and of EBM in particular, is currently experiencing considerable growth. As the technologies are developed and improved, a wider range of industries are beginning to study and implement them to produce an increasingly diverse range of products. High-integrity applications, such as aerospace structural and engine components, stand to become a promising area for the deployment of EBM technology.

However, development in these areas is slowed by the lack of fundamental knowledge, consistent databases, and standardizations, which are all critical in these industry sectors [3-7].

The working principle of EBM consists of consecutive cycles where parts are fabricated layer by layer. An electron beam emitted from a tungsten source is used to selectively melt the component profile on the surface of a metallic powder bed, before a new layer of powder is dispensed over the top and the process is repeated. The electron beam is controlled by a pair of magnetic coils, which focus and control the position, size, and shape of the beam. Manufacturing parameters are generated and controlled by software, which when correctly used, will result in the fabrication of sound parts with high mechanical properties, low porosity and surface roughness, and optimized geometrical reproducibility. The main parameters controlled by the software include: minimum and maximum beam current, number of times the beam scan is to be repeated, scanning speed of the electron beam, distance between individual scan lines (line offset), line order for the hatch pattern, and rotation angle between consecutive hatches [8]. The microstructure evolution and mechanical properties have been studied for conventionally processed Ti-6Al-4V due to its versatility resulting from the good balance between mechanical properties, castability, plastic workability, heat treatability, and weldability [9]. Ti-6Al-4V alloys made by AM have been broadly applied in industry and studied in the laboratory, resulting in an extensive knowledge base compared to other metal alloys fabricated by this technology. More specific studies about the microstructures and static mechanical properties of EBM Ti-6Al-4V were also carried out following the increase in popularity of AM technologies [10-20].

For aircraft and ground vehicle applications, where the AM technologies have a great potential, data that can support accurate design of structural components from a given material, primarily fatigue properties, are necessary [21]. Due to the novelty of AM technologies in general, a limited literature exists on the fatigue performance of AM alloys, and the existing literature is mainly limited to few alloys, of which one of the most studied is Ti-6Al-4V. The major fatigue influence variables studied in the literature are similar to those for traditional cast and wrought products: surface finish, residual stress, internal defects, and microstructure [22]. In this work, fatigue crack growth (FCG) propagation mechanisms at the microstructural scale of EBM Ti-6Al-4V Extra Low Interstitial (ELI) were studied for different stress ratios and growth stages. In a previous study [23], the anisotropic behavior of the material fabricated by EBM was recognized. In that work, FCG properties at low stress ratio ($R=0.1$) were compared between mill-annealed, EBM, and Laser Engineered Net Shape (LENS) Ti-6Al-4V, showing considerable differences in properties depending on the relationship between loading and manufacturing directions. In the same work, solutionized and aged EBM Ti-6Al-4V ELI material was also studied, and differences compared with the as-fabricated condition were observed. Thus, anisotropic behavior is expected in this study due to the directionality of the as-fabricated microstructure, and consequently, FCG results for different stress ratios and directions were compared to those of an equiaxed β annealed microstructure obtained by annealing the material above the β transus temperature.

The relation between microstructure and FCG has been rarely addressed, and a few studies are discussed here. Joshi et al. [24] studied the influence of the electron beam orientation with respect to the crack propagation direction. They concluded that the fatigue resistance of EBM Ti-6Al-4V ELI alloy is greatest when the electron beam orientation is perpendicular to the direction of crack propagation and is consistent between various batches. Seifi et al. [25] studied FCG in five different directions, and reported orientation dependent properties, but a clear relation between the microstructure and the

anisotropic properties was not established. Morton et al. [26] compared the fatigue performance of a wrought alloy with three different microstructures obtained in-situ in the EBM machine through the addition of extra melting stages to the standard manufacturing settings. The FCG tests in direction and horizontal orientations showed that the standard EBM microstructure (with the lowest α lath thickness) had the lowest low cycle fatigue (LCF) fatigue life in both directions. As α lath thickness was increased, the LCF fatigue life was also improved, meeting the wrought alloy properties. However, it should be noted that in the wrought alloy, the enhanced durability of the parts was due to longer crack initiation life; FCG rates being actually higher than those in the EBM material. The standard EBM microstructure in the horizontal direction had the lowest crack propagation rate, but also lowest overall LCF life compared to the vertical direction. Other parameters, like defects and surface roughness were also studied. Edwards et al. [11] evaluated the fatigue performance of Ti-6Al-4V alloy by FCG and high cycle fatigue (HCF) testing of lab specimens, as well as a specific fatigue testing of a bracket component. According to their results, fracture toughness and FCG were comparable to literature properties for wrought Ti-6Al-4V, while the HCF fatigue life was found to be significantly lower than handbook values for wrought Ti-6Al-4V. The rough surface finish in the as-fabricated condition, and porosity in the machined condition led to premature failures. Svensson et al. [18] studied the fatigue performance (axial HCF and rotating beam fatigue) for EBM Ti-6Al-4V, reporting that the micro-porosity attributed to the entrapped argon gas has a negative influence on these properties. After the elimination of the micro-porosity via hot isostatic pressing (HIP) treatment, the fatigue performance was similar to the wrought alloy. The influence of specimen orientation and HIP have also been studied by Facchini et al. [12], who considered the possibility that coarsening of the microstructure due to HIP has a major influence on the increased fatigue resistance, more than the effect of the porosity elimination. This could be understood if the fatigue resistance of the material is considered mainly controlled by crack propagation, where microstructure coarsening has a crack decelerating effect.

Other studies were focused on EBM's surface finish effects on fatigue compared to other AM technologies. Chan et al. [27] investigated the effect of surface roughness on the fatigue life of EBM and LBM (Laser Beam Melting) Ti-6Al-4V using Arcam and EOS technologies, respectively. In this work, three point bend tests were performed with a maximum surface stress of 600 MPa and $R=0.1$ stress ratio. It was found that the fatigue life of rolled Ti-6Al-4V was higher than that of LBM, which in turn was higher than that of the EBM fabricated material. They correlated fatigue life to surface finish, and from their data, an equation linking both parameters was derived. Cronskär et al. [28] deduced that there are clear indicators of considerable reduction of fatigue strength using the as-fabricated EBM surface. Rafi et al. [16] compared the fatigue behavior of Ti-6Al-4V alloy fabricated with EBM and selective laser melting (SLM). Their study reveals a 550 MPa fatigue limit for the samples fabricated with SLM and 340 MPa for those fabricated by EBM. The reason for the better performance of the SLM samples is attributed to the martensitic phase, which impedes dislocation motion, and thus leads to a strengthening effect. The lower fatigue strength of EBM was attributed to its lamellar microstructure rather than the presence of micro-pores or voids.

There has been a renewed interest in the mechanisms responsible for the influence of the stress ratio, R , and the maximum stress intensity, K_{max} , on the FCG threshold, ΔK_{th} [29]. However, studies of AM Ti-6Al-4V under high R ratio conditions and effects on ΔK_{th} and FCG rates have not been widely reported [30]. Traditionally, the differences in ΔK_{th} for different stress ratios are attributed to the presence of crack closure. Several mechanisms of crack closure have been reported in the literature, including

plasticity-, oxide-, and roughness-induced closure. However, previous studies have shown that for high stress ratios ($R > 0.5$) where global crack closure was mostly absent, ΔK_{th} has still decreased with increasing R , indicating that closure was not the only mechanism affecting FCG in early stages [30]. In the absence of closure, possible alternative explanations for this behavior include the presence of sustained-load cracking (SLC), such as creep or hydride-assisted cracking, or the occurrence of the so-called "Marci effect" where at high K_{max} levels, the fatigue threshold ceases to exist, and at all applied ΔK levels (including static loading, where $\Delta K = 0$), the crack advances with a substantial velocity. Both of these phenomena have been observed in titanium alloys at ambient temperatures [29].

The main goal of this study is to systematically investigate the relationships between fatigue crack propagation and microstructure of EBM Ti-6Al-4V ELI, while other factors such as surface roughness and defects have been eliminated by machining the specimens and considering that porosity has limited effect on overall FCG rates. Residual stress that significantly affects the fatigue performance of a material was considered to have a negligible effect for EBM parts. The reason for this assumption is based on the special thermal conditions of this process, resulting in limited residual stress and primarily on the surface of the parts [11]. Thus, machining away the surface, practically eliminates any possible residual stress. Furthermore, the effect of the stress ratio on ΔK_{th} is also addressed, and analytical correlations for vertical and horizontal crack propagation are proposed. The FCG microstructural mechanisms in all regions of crack growth were established for both as-fabricated and heat treated materials in both propagation directions.

2 Materials fabrication and evaluation methods

2.1 Materials and processing conditions

The samples investigated in this study came from a batch manufactured from Ti-6Al-4V ELI gas atomized powder provided by Arcam. Ti-6Al-4V ELI contains reduced levels of oxygen, nitrogen, carbon, and iron (Table 1). The powders used in this study, shown in Figure 1 [19], have a particle diameter between 45 and 150 μm . The powder is made up of 50% new and 50% reused powder. Reused powder is non-sintered powder, which is kept in the machine under vacuum atmosphere for the next build. When the powder is reused, it is possible for the oxygen content to increase, affecting the resulting microstructure and mechanical properties [31]. Nandwana et al. [32] demonstrated that for Ti-6Al-4V powders exposed to the electron beam for 150 hours, the oxygen concentration increases from 0.141 to 0.168 wt%, which is still within the specification for this alloy. In that study, no fresh powder was added such that the worst case scenario was examined. Thus, it is safe to assume that in the current study the oxygen content is within specification since fresh powder was mixed and the starting oxygen concentration for the ELI powder is lower than that for regular Ti-6Al-4V. The EBM manufacturing parameters (spot size, scan velocity, sintering route, etc.) were defined by the internal algorithm of the Arcam Q10 machine, which was running software version 4.2.49 for the build used in this study. The build was started after attaining a preheat temperature of 650°C, each scanning layer was 50 μm thick, and the average scanning time per layer was one minute.

Table 1. Standard chemical composition [33] and actual chemical composition of the powder provided by Arcam [34]

	Al	V	C	Fe	O	N	H	Ti
Ti-6Al-4V ELI (Standard)	5.5-6.5%	3.5-4.5%	<0.08%	<0.25%	<0.13%	<0.05%	<0.012%	Balance

Ti-6Al-4V ELI (Arcam)	6.0%	4.0%	0.03%	0.1%	0.1%	0.01%	<0.003%	Balance
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Figure 1. Ti-6Al-4V ELI powder particles used to fabricate the EBM samples in this study [19].

Beta annealing heat treatment was carried out using a Thermo Scientific Lindberg/Blue (1.8 kW) furnace. The specimens were placed in an argon protected atmosphere furnace at 1100°C for 30 minutes, which was considered enough time for the complete transformation of the microstructure in the whole specimen with a thickness of 15 mm. This temperature is 105°C above to β transus temperature ($\sim 995^\circ\text{C}$), ensuring that at the annealing temperature the microstructure of the sample is 100% β phase while preventing extensive grain growth. After the annealing treatment, specimens were furnace-cooled with a recorded cooling rate of 0.06°C/sec.

The static mechanical properties for as-fabricated and β annealed EBM Ti-6Al-4V ELI were established in previous studies [19,20], and are summarized in Table 2. The static mechanical properties of the β annealed material are not dependent on the loading orientation, so an average value is shown.

Table 2. Static mechanical properties of as-fabricated and β annealed EBM Ti-6Al-4V ELI [19,20]

	Orientation	YS [MPa]	UTS [MPa]	Elongation [%]	Hardness [HV]
As-fabricated EBM Ti-6Al-4V ELI	Vertical	1001 (± 25)	1073 (± 28)	11 (± 2)	366 (± 5)
	Horizontal	1006 (± 29)	1066 (± 10)	15 (± 2)	
β annealed EBM Ti-6Al-4V ELI	Vertical	774 (± 112)	913 (± 38)	13 (± 2)	378 (± 44)
	Horizontal				

2.2 Specimen locations and preparation

The position and orientation of the specimens, as well as the reference axis system on the build platform of the Arcam machine are shown in Figure 2(a). It is important to define the orientation of the samples and establish the terminology used in this study. Other authors have defined up to five different crack propagation directions in the build for FCG experiments [25]. Only two FCG directions (vertical and horizontal) were considered relevant in this study. Build orientation directional differences between the two horizontal directions, X and Y, of the build chamber were neglected as during the building process the electron beam traverse direction rotates by a certain angle with respect to the previous beam traversing direction after every layer [11]. The terms *vertical* and *horizontal* orientations are used to identify crack growth directions that are parallel and perpendicular to the build direction/Z axis. Compact tension, C(T), specimens employed in this study were machined from rectangular blocks

(105 mm x 105 mm x 15mm) located parallel to the four edges of the build platform, with their longitudinal axis parallel to the build direction. The C(T) specimens were machined in different orientations in order to study the crack propagation in vertical and horizontal directions.

Figure 2. (a) Layout of the specimens on the build platform with the standard AM axis system [35] (horizontal [red] and vertical [green] C(T) sample extraction locations are shown); (b) Geometry and dimensions of the C(T) specimens used for the FCG studies.

The Geometry of the C(T) specimens used for the FCG experiments are shown in Figure 2(b) with 35.6 mm (1.4 in.) width (W) and 7.6 mm (0.3 in.) thickness (B) machined in-house using a HAAS CNC mill according to the ASTM-E647 standard [36]. Overall C(T) specimen dimensions were 45 mm x 43 mm x 7.6 mm. The initial notch length (measured from the front face) was 18 mm (0.7 in.). The notch was introduced using wire-cut electrical discharge machining (EDM), and its thickness was 0.254 mm (0.01 in.).

2.3 Testing and analysis

2.3.1 Fatigue crack growth testing

Constant stress ratio tests at $R=0.1$, 0.5 , and 0.8 were performed in room temperature air, $22-24^{\circ}\text{C}$ ($71-75^{\circ}\text{F}$), and relative humidity of $20-50\%$. These tests were run under K -control starting at an initial $K_{\max}$ that would provide a crack growth rate of $5-9 \times 10^{-8}$ mm/cycle ($\sim 10 \text{ MPa}\sqrt{\text{m}}$ for the $R=0.1$ tests, $\sim 13 \text{ MPa}\sqrt{\text{m}}$ for the $R=0.5$ tests, and $\sim 27 \text{ MPa}\sqrt{\text{m}}$ for the $R=0.8$ tests). A cyclic frequency of 20 Hz was used in order to generate FCG data in Regions I and II. A K -gradient, C , of $-0.19/\text{mm}$ was used for the decreasing K part of the test (Region I) to determine the crack growth threshold value, ΔK_{th} , according to the following equations:

$$\Delta K_{\text{app}} = \Delta K_0 \times e^{[C(a-a_0)]} \quad (1)$$

$$C = \frac{1}{K} \times \frac{dK}{da} \quad (2)$$

Region II data were generated using increasing K with a K -gradient of $+0.19/\text{mm}$ at a frequency of 20 Hz . An initial $K_{\max}$ was selected to provide a crack growth rate that was a factor of 5 lower than the initial

crack growth rate observed during the load shedding part of the test, in order to ensure that there were no gaps in the final FCG curve. The final part of the tests was run at constant load (95% of the maximum load reached during the second part to the test) at a cyclic frequency of 5 Hz to obtain data in Region III of fast crack growth. The generated FCG data were analyzed, and fracture toughness values were obtained using Fracture Technology Associates' automated analysis software [37]. Closure evaluations were done using the adjusted compliance ratio (ACR) technique [38,39]. To account for closure effects on FCG, ASTM standard E647 introduced a procedure for using load displacement records to determine the crack opening load under cyclic loading. The method identifies the upper linear portion of the load-displacement record within a defined degree of accuracy, which was regarded as the load range for the effective stress intensity [39]. The ACR method also relies on load-displacement records and calculates the compliance (or displacement/strain range) ratio, using displacement measurements from the front face of the C(T) specimen.

2.3.2 Microstructural and fracture analysis

Metallographic specimens were cut with an abrasive wheel, mechanically ground, polished with colloidal suspension, and etched with Kroll's reagent (2% HF, 6% HNO₃, and 92% DI H₂O). Microstructures were characterized using standard metallographic methods using a Nikon MA 200 Eclipse microscope and image analysis with Elements-D software. To establish the fatigue crack propagation mechanisms, the fracture surfaces were examined using a JEOL-7000F scanning electron microscope (SEM), and subsequently mounted and polished for 2D examination of the crack path along the median section.

3 Results and discussion

3.1 Microstructure characterization of the as-fabricated and β annealed EBM Ti-6Al-4V

The microstructure of the as-fabricated EBM Ti-6Al-4V ELI was characterized in a previous study [19]. It consists of prior β columnar grains oriented parallel to the build direction (Z or vertical axis), Figure 3, with a α + β lamellar structure inside the β grains. The structure of the lamellae is mainly Widmanstätten, or "basket weave", organized into discrete groups of α lamellae with similar thicknesses, referred to in this study as "regions", examples of which are highlighted in Figure 3(e). Although these regions are not often clearly delimited, this concept was employed for the FCG analysis discussed in this study as the size of the columnar grains is not quantifiable due to the difficulty of grain boundary identification. However, some of the prior β grain boundaries are marked by an α layer with an average of thickness 2 μ m, Figure 3(d). This microstructure has been extensively described and characterized in previous studies [10,19,40-43]. At low magnification, scanning layers are also visible, Figure 3(b). These layers have an average thickness of 50 μ m, in agreement with the manufacturing parameters. Due to the anisotropy of the described microstructure with prior β columnar grains oriented in Z direction and scanning layers in XY planes, a different performance for different load orientations and crack propagations is expected.

Figure 3. As-fabricated condition microstructure. (a) 3D microstructural cube (Z axis corresponds to the build direction); (b-e) XZ plane microstructure at 5x, 100x, 500x, and 1000x magnifications. An example of an α layer at the prior β grain boundaries is shown in (d); examples of fine (red) and coarse (yellow) α regions are circled in (e) [19].

The microstructure obtained after β annealing and slow cooling in the $\alpha+\beta$ field, is $\alpha+\beta$ lamellar with large, clearly defined, colonies of aligned α plates in prior β equiaxed grains, and coarse α layers at grain boundaries. Because of annealing above the β transus temperature, the morphology of the β grains changes from columnar, as visible in Figure 3(c), to equiaxed, Figure 4(a). When comparing the microstructures at high magnification an extensive coarsening of the microstructure produced by the slow furnace-cooling is observed; the measured α lath thickness for the as-fabricated condition, Figure 3(d) is $\sim 0.7\ \mu\text{m}$ while for the β annealed, Figure 4(b) is $\sim 4.4\ \mu\text{m}$. The $\alpha+\beta$ lamellae in the β annealed condition form colonies while the as-fabricated microstructure mainly consists of “basket weave” morphology. Isotropic properties are expected for the β annealed condition because the directionality of the as- fabricated condition was eliminated, becoming an equiaxed structure.

Figure 4. XY plane microstructure of the β annealed condition after 30 minutes at 1100°C in the (a) 100x and (b) 500x magnifications; an example of an α colony is outlined in orange in (a).

3.2 Fatigue crack growth data and mechanisms in the near threshold regime

3.2.1 Effect of the stress ratio

Fatigue crack propagation data were generated for a range of stress ratios for both as-fabricated, Figures 5(a,b), and β annealed, Figures 5(c,d), conditions in vertical and horizontal orientations.

Figure 5. Fatigue crack growth behavior of (a,b) as-fabricated and (c,d) β annealed materials in vertical and horizontal orientations for different stress ratios.

An expected R-ratio effect on the FCG rate was observed in the near-threshold regime even at high ratios, and especially for the as-fabricated horizontal crack propagation. However, as ΔK increased in the middle-upper Region II, the influence of R on da/dN decreased. Additionally, the material does not exhibit the “Marci effect” previously described. With increasing stress ratio, a decrease in the ΔK_{th} and increase in the near-threshold FCG rates were observed for all conditions, as shown graphically in Figure 6(a). For the as-fabricated condition, it can be observed that ΔK_{th} in the vertical direction is higher than in the horizontal direction for all stress ratios tested. The β annealed material exhibits a considerably higher ΔK_{th} than the as-fabricated material at all stress ratios. However, for this material there is no appreciable difference between ΔK_{th} in the horizontal and vertical orientations, which was expected due to the equiaxed heat treated microstructure.

The higher ΔK_{th} seen at low stress ratios is commonly related to crack closure effects [44]. Values for the applied ΔK_{th} and ΔK_{th-eff} are listed in Table 3 (ΔK_{th-eff} is the value of the stress intensity range at threshold after ACR crack closure corrections). ΔK_{th-eff} can also be used to numerically describe closure, and provide a quantitative understanding of its influence for different stress ratios and orientations. The closure ratios calculated using Eq. (3) has been plotted versus stress ratio for different materials and conditions in Figure 6(b).

$$U = 1 - \left(\frac{\Delta K_{th-eff}}{\Delta K_{th}} \right) \quad (3)$$

It can be seen from Figure 6(b) that closure ratio decreases as stress ratio increases since higher stress ratio ΔK_{th} is less affected by closure. It can also be observed that the closure ratio, and therefore closure effects, in the as-fabricated material is lower in the horizontal direction for all stress ratios. This observation indicates that there is a lower roughness-induced closure for the crack propagation in the horizontal direction compared to the vertical direction. In the β annealed condition, the effects of closure are significantly pronounced for a stress ratio of R=0.1, but are similar to the as-fabricated vertical results at R=0.5 and R=0.8. In addition, it can be observed that there is little appreciable closure difference between the β annealed horizontal and vertical orientations, due to the similar crack propagation behavior of the two orientations.

Figure 6. (a) Stress ratio effect on the ΔK_{th} and (b) Closure ratio (U) versus stress ratio (R) for vertical and horizontal crack propagation directions in as-fabricated and β annealed conditions.

Table 3. Values of ΔK_{th} , ΔK_{th-eff} , and closure ratios for different stress ratios and crack propagation directions in as-fabricated and β annealed conditions

Material Condition	Stress Ratio R	ΔK_{th} [MPa $\sqrt{m}$]		ΔK_{th-eff} [MPa $\sqrt{m}$]		$U=[1-(\Delta K_{th-eff}/\Delta K_{th})]$	
		Vertical	Horizontal	Vertical	Horizontal	Vertical	Horizontal
As-fabricated	0.1	4.2	3.7	3.6	3.2	0.14	0.13
	0.5	3.0	2.7	2.7	2.5	0.10	0.07
	0.8	2.9	2.2	2.6	2.1	0.10	0.04
β Annealed	0.1	6.0	6.0	3.9	3.8	0.35	0.37
	0.5	-	4.8	-	4.3	-	0.10
	0.8	4.6	4.5	4.0	4.0	0.12	0.11

3.2.2 Threshold-stress ratio correlations

To predict FCG rates properly, particularly in early growth stages, an accurate evaluation of the threshold is necessary. One challenge in comparing data at different stress ratios is incorporating the effects of the differences in both the mean and maximum stress levels [22]. In order to facilitate comparisons with existing and future fatigue studies, especially in the near-threshold region, a procedure for correlating ΔK_{th} and stress ratio is presented in this study following a method proposed by Döker [45]. In this method, the FCG behavior of a material in the near-threshold regime is described by four constants, which can be readily determined. This method was employed successfully for conventionally processed Ti-6Al-4V by other authors [29,30].

Using the data obtained from FCG tests at different stress ratios and directions, Figure 5, the $K_{max,th}$ -R and ΔK_{th} - K_{max} diagrams for the as-fabricated and β annealed EBM Ti-6Al-4V ELI alloy were developed and are presented in Figures 7 (note that only the horizontal β annealed case is examined, since there is no appreciable difference between the two orientations). According to Döker, the $K_{max,th}$ -R curve at low R ratios and the ΔK_{th} - K_{max} curve for high K_{max} values can be represented by two straight lines, with the mathematical expressions in Eqs. (4,5):

$$K_{max,th} = (K_{max,0} + \beta R) \quad (4)$$

$$\Delta K_{th} = \Delta K_0 + \alpha K_{max} \quad (5)$$

where $K_{max,0}$ and ΔK_0 are the intersection points of two straight lines with the vertical axes, and α and β are the respective slopes. These values are derived from the plots in Figure 7, and are shown in Table 4 for both vertical and horizontal directions in the as-fabricated material, as well as the horizontal direction in the β annealed material. ΔK_0 is proportional to the elastic modulus of the material, and $K_{max,0}$ is influenced by microstructure and environment.

Table 4. Values for intersection points and slopes for as-fabricated and β annealed EBM Ti-6Al-4V ELI in vertical and horizontal directions

	As-fabricated		β annealed
	Vertical	Horizontal	Horizontal
$K_{max,0}$ [MPa $\sqrt{m}$]	4.4	3.8	5.9
β [MPa $\sqrt{m}$]	3.45	3.50	6.75
ΔK_0 [MPa $\sqrt{m}$]	3.2	3.3	4.7
α [MPa $\sqrt{m}$]	-0.02	-0.10	-0.01

Figure 7(a) $K_{max,th}$ -R and (b) ΔK_{th} - K_{max} diagrams of EBM Ti-6Al-4V from where α , β , ΔK_0 , and $K_{max,0}$ were calculated.

Figure 7(b) is relevant to fatigue-based design because it shows graphically when crack growth will occur (i.e. for conditions above the curve). For as fabricated material, it can be seen that for the horizontal crack growth the area above the curve is larger than for the vertical crack growth, which means that cracks can propagate at lower K_{max} and ΔK_{th} values in the horizontal direction. The area above the β annealed data points is even smaller, as the material is more resistant to crack growth. This can be used as a material optimization tool, identifying processing or heat treating conditions that reduce the area above the ΔK_{th} - K_{max} curve.

Applying Eqs. (6,7) to the threshold, the final ΔK_{th} -R correlations were developed as given in Eqs. (8,9):

$$R = \frac{K_{min}}{K_{max}} \quad (6)$$

$$\Delta K = K_{max} - K_{min} \quad (7)$$

$$\Delta K_{th} = (K_{max,0} + \beta R)(1 - R) \quad \text{for } R \leq R_c \quad (8)$$

$$\Delta K_{th} = \left(\frac{1 - R}{1 - R - \alpha} \right) \Delta K_0 \quad \text{for } R > R_c \quad (9)$$

where R_c represents the critical stress ratio where the curves generated from Eqs. (8,9) intersect. The physical meaning is that for stress ratios above the critical value, roughness-induced closure is minimal. The calculations in this study have been made using the data generated from only three stress ratios ($R=0.1, 0.5$, and 0.8), therefore, the arbitrary value for the critical stress ratios is $R_c=0.5$. This value shows an apparent transition from “closure-affected” to “closure-free” thresholds, and is supported by data obtained in previous studies [29,30], as well as the noticeable slope change in the ΔK_{th} - K_{max} diagram, Figure 7(b). The graphical representation of the ΔK_{th} - R correlations for stress ratios below and above R_c are shown in Figure 8. The ΔK_{th} - R curves are useful tools for estimating the threshold behavior for conditions and stress ratios that have not been experimentally studied which is particularly useful for studies where specimen numbers are limited.

Figure 8. Calculated ΔK_{th} - R correlations for vertical and horizontal crack propagation (dotted lines), together with the experimental results.

3.2.3 Effects of the microstructure on fatigue crack growth threshold

3.2.3.1 Fatigue crack growth observations

Figure 9 shows the FCG plots for both horizontal and vertical crack propagation in the as-fabricated and β annealed conditions, at stress ratios of $R=0.1$ and 0.8 . The $R=0.5$ results have not been included in the following discussions as the comparisons between $R=0.1$ and 0.8 are sufficient for clarifying the mechanisms at the microstructural scale and highlight the differences between orientations as well as between the as-fabricated and heat treated materials. The data in Figure 9 indicate that for the β annealed conditions, FCG behavior in both orientations is identical at both stress ratios as expected due to the equiaxed microstructure. Comparing the results to the as-fabricated and β annealed conditions, a considerable improvement of the FCG resistance of the β annealed material is observed for both stress ratios and at all growth stages.

Figure 9. FCG of as-fabricated and β annealed Ti-6Al-4V ELI in horizontal and vertical directions tested at: (a) $R=0.1$ and (b) $R=0.8$.

The improvement of the FCG properties of the β annealed microstructure, compared to the as-fabricated microstructure, can be attributed to the higher FCG resistance of a coarse lamellar structure versus a fine lamellar structure. The FCG resistance of a lamellar microstructure in the near -threshold regime is influenced by three main factors: crack closure, the intrinsic material properties, and crack front geometry, with crack closure being the dominating effect at low stress ratios, and crack front geometry controlling the FCG behavior at higher stress ratios. All three mechanisms are dependent on the α colony/region size of the material; in this study, the as-fabricated material has a much finer α colony size than the β annealed material. As the α colony size in a lamellar material increases, the influence of crack closure and crack front geometry also increase, raising the FCG resistance of the material in early stages. It should be noted that the intrinsic material resistance to crack growth (i.e., ductility) *decreases* as the α colony size increases, but its influence is overridden by the increasing effect of the other two factors [46].

Thus, at low stress ratios, the increase in the near-threshold FCG resistance of the β annealed materials, is explained by higher roughness-induced closure. This statement can be corroborated by examining the applied, and closure-corrected, threshold values (ΔK_{th} and ΔK_{th-eff} , respectively) in Table 3. The closure corrected threshold of the β annealed material, ΔK_{th-eff} , is similar with the as-fabricated vertical and horizontal values. The greater influence of closure in the β annealed condition is also observed by the considerably higher closure ratio, $U=0.39$ and 0.37 , compared to the as-fabricated values of $U=0.16$ and 0.13 .

Comparatively, at high stress ratios the effect of closure is reduced due to the higher minimum load, which may be expected to diminish the magnitude of the increase in the FCG resistance observed in the β annealed material [47]. However, the increase in the ΔK_{th} of the β annealed microstructure compared to the as-fabricated one is the same at both stress ratios; an increase of $1.7 \text{ MPa}\sqrt{\text{m}}$ in the vertical orientation and $2.3 \text{ MPa}\sqrt{\text{m}}$ in the horizontal orientation are observed. The effects of oxidation, which can enhance closure effects at higher stress ratios, can be ruled out as a contributing factor since oxide

layers in Ti-6Al-4V do not significantly increase in thickness at temperatures under 200°C [48]. It follows that the improved FCG resistance of the β annealed material at high stress ratios is related to the effects of crack front geometry. This phenomenon is explained by considering two adjacent areas ahead of the crack front with different crystallographic orientations, which can cause the crack front to move locally out of the main propagation plane in directions determined by the different slip planes in the two adjacent colonies. This mechanism is shown schematically in Figure 10. Thus, the crack locally deviates from the average propagation plane over a distance which is proportional to the size of the characteristic microstructural features. Consequently, the crack front becomes bifurcated and the local crack tips in the two adjacent areas are separated by a distance Z perpendicular to the main propagation direction, Figure 10. In the context of this study, the distance Z is proportional to the α colony size of the material. In order to move this bifurcated crack front forward, the distance Z has to be traversed by propagating the crack in an unfavorable direction along unfavorably oriented crystallographic planes. The resistance of this crack propagation obstacle is directly proportional to the distance Z . Thus, the strength of the obstacle is larger for coarse microstructures than for fine microstructures, resulting in better crack propagation resistance in materials with larger microstructural colony sizes [46]. With reference to this study, this means that the β annealed material (coarse α colony size) is expected to have greater FCG resistance than the as-fabricated material (fine α colony size), as observed in Figure 9.

Figure 10. Schematic local crack front configuration with two adjacent colonies A and B having different crystallographic orientations, and therefore differently oriented major slip/crack planes; image reproduced from reference [46].

3.2.3.2 Fractography and microstructural mechanisms at threshold

Figures 11 and 12 show the crack paths and fracture surfaces of the as-fabricated material in the near-threshold region for two stress ratios, $R=0.1$ and 0.8 . Figure 11 shows the vertical crack propagation and Figure 12 the horizontal crack propagation.

As-fabricated, near threshold, vertical crack propagation		
$R=0.1$ $\Delta K_{th} = 4.3 \text{ MPa}\sqrt{\text{m}}$	<p>SEM image showing the crack path for $R=0.1$. The crack is dark and jagged, propagating horizontally. A scale bar of 50 μm is visible in the bottom right corner.</p>	<p>SEM image showing the fracture surface for $R=0.1$. The surface is rough and textured. Technical data at the bottom: WPI 3 SEM SEI 10.0kV X500 WD 14.7mm 10um.</p>
$R=0.8$ $\Delta K_{th} = 2.9 \text{ MPa}\sqrt{\text{m}}$	<p>SEM image showing the crack path for $R=0.8$. The crack is dark and relatively smooth, propagating horizontally. A scale bar of 50 μm is visible in the bottom right corner.</p>	<p>SEM image showing the fracture surface for $R=0.8$. The surface is rough and textured. Technical data at the bottom: WPI 3 SEM SEI 10.0kV X500 WD 14.7mm 10um.</p>

Figure 11. Near-threshold crack paths and fracture surfaces for vertical crack propagation in the as-fabricated material for two stress ratios, $R=0.1$ (top) and $R=0.8$ (bottom).

Figure 12. Near-threshold crack paths and fracture surfaces for horizontal crack propagation in the as-fabricated material for two stress ratios, $R=0.1$ (top) and $R=0.8$ (bottom).

The comparison of the crack profiles at different stress ratios shows a significantly higher amount of branching and secondary cracking for $R=0.1$ in both vertical and horizontal directions. This behavior is more accentuated in the vertical crack propagation. From the microscopy work, it was observed that the crack has a higher tendency to form steps where it interacts with coarser α lath regions, while showing a more continuous path through the finer α lath regions. The preferential orientation of the α lath in the “basket weave” also has an influence on the crack path, offering a favorable path when the direction of the laths are close to the direction of the crack. Some cracks also exhibit growth along region boundaries, but because of the complexity of the microstructure, the quantification of the propagation associated with each feature is challenging. This observation is corroborated by studying the associated fracture surfaces, where longitudinal features following the crack propagation direction have been observed for all stress ratios, but only in the vertical direction. It is thought that these longitudinal features are prior β grain boundaries comprised of α phase as shown in Figure 3(b). The effects of crack branching and tortuosity can be evaluated from an energy-dissipation perspective; the more tortuous the crack, the more energy required to extend it along the nominal crack plane. Furthermore, the measured nominal crack growth rate (relative to the sample axes) will be less than the actual rate once a new crack surface is formed [49]. Other studies conducted on SLM Ti-6Al-4V also detected lower FCG rates in the crack propagation direction parallel to the columnar grains [50,51]. However, the microstructure, although similar to EBM and its fine features, has some differences, including the presence of α' martensite, which is almost inexistent in the EBM as-fabricated material.

A study of the 2D crack paths and the fracture surfaces for the β annealed condition was also performed. Figure 13 shows the vertical crack propagation in β annealed material at stress ratios of $R=0.1$ and 0.8 . Only the vertical orientation is shown for the β annealed condition since it has already been established in Section 3.1 that the heat treated microstructure is equiaxed and orientation no longer affects the crack propagation behavior.

Figure 13. Near-threshold crack paths and fracture surfaces for vertical crack propagation in the β annealed material for two stress ratios, $R=0.1$ (top) and $R=0.8$ (bottom).

Looking at the crack fronts in a plane perpendicular to the crack propagation direction, the large difference between the two microstructures is evident. The crack front profile for the β annealed microstructure is rough, and the steps in the Z direction, caused by interaction with α colonies, are clearly visible, and lie in an unfavorable orientation for crack propagation, Figure 13. By comparison, the crack front profiles of the as-fabricated microstructures are smoother because the height of the Z steps are smaller, as observed in Figures 11 and 12.

Analyzing the relative orientation of the crack propagation and microstructural directionality in the as-fabricated condition, it can be seen that the horizontal crack propagation is perpendicular to the columnar grains and parallel to the scanning layers. Therefore, the horizontal crack will cross the α phase at grain boundaries at angles close to 90° . On the other hand, the crack propagation in the vertical direction occurs parallel to the grain boundaries and perpendicular to the scanning layers, as shown in Figure 14. In the as-fabricated condition, the difference between ΔK_{th} values in the near-threshold region, Figure 15, suggests characteristic interactions between the directional microstructure and the fatigue cracks.

Figure 14. Schematic representation of the crack propagation in vertical and horizontal directions

Figure 15. Detail of the FCG curve in the near-threshold region for 0.1 and 0.8 stress ratios for vertical and horizontal crack propagation in the as-fabricated material.

Considering the characteristics of the crack paths in vertical and horizontal directions for the as-fabricated conditions, together with the relative orientations between the crack propagation direction and prior β columnar grains and scanning layers, the crack interaction mechanisms at the microstructure scale were established. For crack growth in the horizontal direction, the prior β grain boundaries act as barriers to the crack growth; the horizontal crack propagates in the same direction as the scanning layers and cuts directly through the prior β grain boundaries without considerable interaction with them as shown in Figures 16(a). For crack growth in the vertical direction, the prior β grain boundaries act as crack deflectors in angles close to their orientation, interacting with the crack and decreasing the crack propagation rate in the vertical direction. The columnar grain structure is responsible for the branching of the crack, while the scanning layers contribute to the deflection of the crack front enhancing the tortuosity of the crack path as shown schematically in Figure 16(b).

(a)

(b)

Figure 16. Schematic view of the crack propagation in as-fabricated material in (a) horizontal and (b) vertical directions, showing the interactions with prior β columnar grain boundaries and scanning layers.

The FCG resistance of β annealed Ti-6Al-4V is often attributed to large amounts of crack branching (when the crack splits or two or more crack tips advance simultaneously) and tortuosity (large-scale meandering of the crack from its nominal plane, defined by the notch). The relative orientation of the lamellae at the crack tip determines, to a large extent, the path along which the crack propagates. If the long axes of the lamellae are oriented relatively parallel to the load axis, the crack will cut directly across the lamellae while the crack growth along the lamellae boundaries is preferred when the lamellae are aligned perpendicular to the load axis [49].

3.3 Fatigue crack growth microstructural mechanisms in Regions II and III

The estimated fracture toughness of the as-fabricated material is higher for the horizontal direction than the vertical direction ($56.6 \text{ MPa}\sqrt{\text{m}}$ compared to $49.7 \text{ MPa}\sqrt{\text{m}}$ at $R=0.1$), similar to the observations by Edwards et al. [11]. The fracture toughness also tends to decrease with increasing stress ratio for all tested conditions. For example, the vertically orientated as-fabricated specimens, Figure 5(a), show a reduction in fracture toughness from $56.6 \text{ MPa}\sqrt{\text{m}}$ at $R=0.1$, to $28.8 \text{ MPa}\sqrt{\text{m}}$ at $R=0.5$, and $17.7 \text{ MPa}\sqrt{\text{m}}$ at $R=0.8$. The β annealed material tends to have a higher fracture toughness, for example vertically orientated as-fabricated and β annealed specimens tested at $R=0.1$ have estimated fracture toughness values of $49.7 \text{ MPa}\sqrt{\text{m}}$ and $82.0 \text{ MPa}\sqrt{\text{m}}$, respectively. The lower yield strength of the β annealed material is a contributing factor to the higher fracture toughness, **but in addition, previous studies of $\alpha+\beta$ titanium alloys have linked the increase of Region II FCG rates and fracture toughness values with increasing α colony size due to the contribution from the rougher crack front geometry dominating over the ductility effects [46].**

Figures 17 and 18 show Regions II & III crack paths and fracture surfaces for vertical and horizontal crack propagation in as-fabricated material, and Figure 19 shows vertical crack propagation in the β annealed material at stress ratios of $R=0.1$ and 0.8 . For the as-fabricated condition it is not easy to identify clear microstructural features, such as grain boundaries or colonies. Therefore, it is necessary to go to higher magnifications to observe the fracture mode and interactions between the crack and microstructure, which is still challenging due to the fine characteristics of the as-fabricated microstructure. Fracture in

the near-threshold region shows considerable branching as it was previously described in Section 3.2.3.2, where crack growth in vertical and horizontal crack propagation directions were described and compared. Region II of crack growth in the as-fabricated material shows a less structure-sensitive fracture mode with a smoother crack topography, mainly due to reduced branching of the crack, Figures 17 and 18. Region III shows the approach of a static fracture mode with a rougher fracture surface. The study of the fracture surface for the β annealed specimen shows an evident change in fracture mode as the crack progresses. The fracture in the near-threshold region, propagates mainly in an intra-granular and intra-colony mode. There are also occurrences of inter-granular and inter-colony fracture mode, being in general a very microstructure sensitive fatigue crack propagation. As the crack progresses into Region II and the crack growth rate increases, the crack growth becomes less sensitive to the main microstructural features (grains and colonies), being a continuum mode fracture, as it can be inferred from the flat topography of the fracture surfaces, Figure 19. The fracture mode changes again in Region III to a “static mode” behavior with dimples and a rougher fracture surface appearance. In the as-fabricated material in both vertical and horizontal orientations, as the stress ratio was increased from $R=0.1$ to 0.8 , Region II of crack growth appears less influenced by the microstructural features, resulting in lower tortuosity and less crack branching, while in Region III fracture surfaces were significantly rougher at higher stress ratios, Figures 17 and 18. Also, fracture surfaces of the horizontal direction for both $R=0.1$ and 0.8 exhibit more ductile fracture corresponding to the higher fracture toughness in this direction. In the β annealed material, increasing the stress ratio caused the crack growth in Regions II and III to become more affected by interactions with colony boundaries, indicated by the meandering behavior of the crack and its step-like features.

As-fabricated, Regions II and III, vertical crack propagation		
R=0.1	<p>Region II $\Delta K = 10 \text{ MPa}\sqrt{\text{m}}$</p>	<p>Region III $\Delta K = 40 \text{ MPa}\sqrt{\text{m}}$</p>
	<p>Region II $\Delta K = 6 \text{ MPa}\sqrt{\text{m}}$</p>	<p>Region III $\Delta K = 15 \text{ MPa}\sqrt{\text{m}}$</p>

Figure 17. Region II & III crack paths and fracture surfaces for vertical crack propagation in as-fabricated material at stress ratios of $R=0.1$ and 0.8 .

Figure 18. Region II & III crack paths and fracture surfaces for horizontal crack propagation in as-fabricated material at stress ratios of $R=0.1$ and 0.8 .

Figure 19. Region II & III crack paths and fracture surfaces for vertical crack propagation in β annealed material at stress ratios of $R=0.1$ and 0.8 .

4 CONCLUSIONS

The results presented in this study lead to the following conclusions:

- FCG threshold values were observed to be on average $\sim 15\%$ higher for the vertical crack propagation compared to the horizontal crack propagation in the as-fabricated EBM Ti-6Al-4V ELI at all tested stress ratios;
- The stress ratio effects on threshold were evaluated using Döcker's method, which is based on a two driving-force model for fatigue and advocates two intrinsic thresholds, $K_{\max,0}$ and ΔK_0 . This method can be used to predict ΔK_{th} values for stress ratios that have not been experimentally measured.
- Comparisons between the applied and effective ΔK_{th} of the as-fabricated EBM Ti-6Al-4V ELI show that crack growth in the vertical orientation has larger roughness-induced closure effects;
- The higher branching induced by the prior β columnar grains, and the tortuosity intensified by the scanning layers, makes the vertical crack propagation more energy demanding and in consequence less likely than the horizontal crack propagation in early stages;
- The threshold behavior of the EBM Ti-6Al-4V ELI material can be considerably improved by a β annealing heat treatment, resulting in a 35-45% increase in ΔK_{th} compared to the as-fabricated material. The coarsening of the microstructure after this treatment eliminates the anisotropy of the as-fabricated condition and raises the irregularity of the crack path which also considerably increases the roughness-induced closure.

- In the as-fabricated material, Region II crack growth showed a less structure-sensitive fracture mode with a smoother crack topography due to reduced branching, while Region III was characterized by a rougher fracture surface. The Regions II and III crack growth in the β annealed material displayed similar trends.
- Fracture surfaces for horizontal crack growth in the as-fabricated material exhibited higher levels of ductility, corresponding to the higher fracture toughness in the horizontal orientation compared to the vertical orientation (56.6 MPa $\sqrt{m}$ compared to 49.7 MPa $\sqrt{m}$ at a stress ratio $R=0.1$). Fracture toughness was also found to increase in the β annealed condition (reaching 82 MPa $\sqrt{m}$ in the vertical orientation at $R=0.1$). This is due to reduced ductility and increased α colony size in the β annealed condition, leading to rougher crack front geometries.
- As the stress ratio was increased, Region II crack growth in the as-fabricated material was less influenced by microstructural features with lower tortuosity and less branching of the crack; while Region III exhibited a significantly rougher fracture surface. The Region II and III growth in the β annealed material produced a more faceted fracture surface as the stress ratio was increased.

High-integrity industrial sectors provide promising areas for the application for EBM technology. However, development in these areas is hindered by the lack of knowledge, consistent databases, and standardizations for AM materials and components. This work presented in this paper provides a much needed fundamental understanding of the FCG behaviour of AM-EBM fabricated materials, as well as important design considerations for high-integrity applications

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