Fatigue crack growth behavior of beta-annealed Ti-6Al-2Sn-4Zr-xMo (x=2, 4 and 6) alloys: Influence of microstructure and stress ratio

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Abstract:

The influence of microstructures (β grain size, α/β colony size and width of α lath) and stress ratios ($R=0.1, 0.7$) on fatigue crack growth (FCG) behavior of beta-annealed Ti624x (x=2, 4 and 6) alloys was investigated. The FCG rate curves, crack paths, and the plastic zone ahead of the crack tip were compared. The reducing FCG resistance from Ti6242 to Ti6246 was found to stem from decreasing α/β colony size. Increasing stress ratio also reduced crack deflection and crack closure, and thus decreased the FCG resistance. The relevant mechanisms were discussed in terms of K_{max} and crack closure effect.

Keywords: *Ti624x, Microstructure, Fatigue Crack Growth, Stress Ratio, Crack Closure*

1 Introduction

Titanium alloys are extensively used in aerospace applications due to their high specific strength (strength/density), attractive high-temperature mechanical properties and excellent resistance to corrosion [1, 2]. Since 1970s, damage tolerance design (including fracture toughness and FCG) has been proposed by aviation industry designers as a new design concept in addition to conventional fatigue / strength based criteria in the component-safety design and material selection [3, 4]. FCG properties of titanium alloys, including near-threshold and Paris region, are of great significance and have been extensively studied by many investigators [5-19]. A good understanding of the FCG behavior is an important precondition for the ‘run on’

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fatigue life assessment of service components and the formulation of the inspection and repair schemes.

The predominant influencing factors on FCG behavior which may be divided into extrinsic parameters, such as loading conditions (frequency [10, 11, 20], stress ratio [8, 12, 21] and loading waveform [22, 23]) and environment (temperature [10, 12] and atmosphere [24, 25]), and intrinsic parameters, such as alloy composition (mainly about the interstitial element [1, 24, 26]) and microstructure [7, 8, 21, 27, 28]. These factors are hard to decouple and influence the final FCG behavior with the observed effects like crack bifurcation and deflection [29], crack closure [30], crack blunting [31] and so on.

One of the main agreed conclusions is that the presence of roughness-induced crack closure (RICC) and associated crack path deflection would reduce the FCG rates especially in the near-threshold region. Many researchers have pointed out that stress ratio (R , ratio of minimum to maximum stress) had a direct relationship with RICC where a higher stress ratio would make fracture surfaces avoid contact. Until now, most of the research related to stress ratio effect (also called mean stress effect) was based on the mathematical expression descriptions about FCG rates [12, 13, 16, 32-35], where stress ratio effect has been taken into account (basically make an amendment to the classic Paris equation [36]). Although crack closure, which induces enhanced shielding effect on FCG, can be rationalized by the effective (ΔK_{eff}) rather than by the applied stress intensity factor range (ΔK) to some extent, the specific mechanism of stress ratio acting on FCG behavior is complicated [8] and still not fully understood.

In addition, most widely reported FCG data of titanium alloys is about fully lamellar or Widmanstätten microstructures. Both of the microstructures are thought to be of good damage-tolerance properties, where the size of α/β colony and α lath size have been taken as the main microstructural parameters to be considered [11, 37]. IMI685 alloy was the first titanium alloy applied with a fully lamellar microstructure in view of its high creep resistance and good damage-tolerance properties [38]. At present, most of the FCG research of titanium alloys is focused on Ti-6Al-4V alloy [7, 12, 30,

37, 39-44]. Limited data about the FCG behavior of high-temperature and high-strength titanium alloys have been reported. This was because most of these alloys, such as Ti-6Al-2Sn-4Zr-2Mo (Ti6242) and Ti-6Al-2Sn-4Zr-6Mo (Ti6246), were applied in the form of forgings [1]. However, with the development of melting and casting technology, titanium alloy castings have become increasingly popular in the aerospace industry due to the benefit of lower cost and improved fatigue life with reduced fastening of different parts [1, 27, 28]. Different cast titanium alloys have been developed according to the specific application requirements, *e.g.*, cast Ti6242 for moderately elevated temperature application [27]. The microstructure of the castings generally consists of extra-coarse prior β grains with lamellar or Widmanstätten microstructures which depend on the thickness of component, casting modes and processing control [28, 45]. However, a detailed understanding of FCG behavior in those high-temperature and high-strength titanium alloys with lamellar microstructures is still lacking.

Therefore, in this present investigation, a comparative study was undertaken to characterize the long crack FCG behavior in relation with the alloy types, different lamellar microstructures and stress ratios with Compact Tension (CT) tests in the case of three $\alpha + \beta$ titanium alloys with nominal composition of Ti-6Al-2Sn-4Zr-xMo (Ti624x, x=2, 4 and 6). Two types of annealing temperatures above the β transus (T_β) have been adopted to produce the lamellar microstructures with normal and extra-coarse prior β grains, for simulating cast microstructures. Test results are analyzed in the light of crack growth paths, stress ratio effects and crack closure mechanisms. Comparison of the fracture paths provides the evidence of the microstructure features and stress ratios co-affecting the final FCG behavior.

2 Materials and Experimental Procedure

2.1 Materials

Ti624x (x=2, 4 and 6) ingots were prepared through three times vacuum arc remelting, and then subjected to β and $\alpha + \beta$ phase field deformation successively. The final products were in the form of round bars with a diameter of 90 mm. The details of the processing route have been described elsewhere [46]. The T_β temperature of Ti624x

(x=2, 4 and 6) alloys were about 987 °C, 967 °C and 937 °C, respectively. Specimens for FCG tests were cut from the forged round bars and heat treated to possess fully lamellar microstructures. The typical heat treatment route was solution annealed above T_{β} for 1 hour and slow cooled with 1°C/min to 700 °C and then air-cooled, followed by aging treatment at 595 °C for 8 hours and air-cooled. Two types of β annealing temperatures were chosen to obtain normal and extra-coarse prior β grains for comparative study: one was $T_{\beta} + 20$ °C for the respective alloy, the other was to make Ti624x alloys have similar average prior β grain size of about 2090 μm (annealing temperatures for the respective alloy were 1240 °C for Ti6242, 1300 °C for Ti6244 and Ti6246. The specific temperatures were obtained through a systematic research of β grain growth behavior [47], where the corresponding mechanical properties have also been tested).

2.2 Fatigue Tests

FCG tests were carried out on CT specimens with a size of 10 mm thick and 40 mm wide, as shown in Figure 1. Using a computer controlled MTS 810 servo-hydraulic fatigue test machine, load-controlling tests with constant amplitude were performed with a sinusoidal waveform and a test frequency of 10 Hz at ambient temperature. Two stress ratios were selected for the comparison, namely 0.1 and 0.7, where the corresponding peak applied forces were 5 and 13 kN, respectively. In this case, it would guarantee the similar initial ΔK value (about 10 $\text{MPa}\sqrt{\text{m}}$). This was a constant force amplitude test, where $\Delta K = K_{\max}(1-R)$, ($K_{\max}$ is the maximum stress intensity factor). And ΔK increased with the increasing of the crack length. The crack length was monitored as a function of cycles using the crack opening displacement (COD) compliance procedure. CT specimens were firstly pre-cracked to a length of $a/W = 0.2$ by using electric discharge machining and then subjected to the fatigue pre-cracking until a crack length (a) of $a/W = 0.25$ was achieved under sinusoidal loading at 10 Hz. For the fatigue pre-crack preparation, the precrack final maximum K value and crack length (10 mm for the current study) were set based on the consideration that the plastic zone sizes at the end of the fatigue pre-crack stage were less than those generated at the beginning of FCG testings. All the tests were conducted in

accordance with ASTM-E647 recommendations [48] and under conditions allowing the application of linear elastic fracture mechanics (LEFM). Therefore, the FCG behavior in current study was characterized for six different lamellar microstructures and two stress ratios. Repetitive experiments have been conducted on Ti6242 alloy and produced similar results, which guaranteed the reliability and repeatability of the experiment results.

2.3 Characterization Methods

Microscopic examinations were carried out using both scanning electron microscopes (SEMs) and optical microscopy (OM). The fully lamellar microstructures have been examined in a FEI Inspect-F50 SEM at an accelerating voltage of 20 kV. The FCG path profiles, obtained by sectioning the FCG samples along the middle plane parallel to the sample broad surface, were observed under an OM (Zeiss-Axiover 200MAT). The interaction of FCG path (initial stage) and local microstructures has been analyzed with the aid of electron back scattering diffraction (EBSD) technique. EBSD examination was completed on a HITACHI S-4300N equipped with Channel 5 HKL detector, where the analysis software is Oxford Aztec. A step size of 5 μm was adopted for the α/β colony crystallographic orientation analysis. Working distance is 15 mm and working voltage is 20 kV.

Quantitative measurements were conducted on each microstructure in order to determine the average diameter of prior β grains and α/β colonies, as well as the thickness of α laths. The prior β grain size was determined using the area-measure function in Image-Pro Plus software (by reading on OM images taken after Kroll's etch), and then make the measured area of each β grain to be equivalent to the area of a circle with a diameter of D . In view of the considerable anisotropy in terms of the size of α/β colonies [37], three circles and linear intercept methods in accordance with the ASTM-E112 recommendations [49], were adopted for the estimation of α/β colony and α lath size, respectively. The linear intercept method was using the line perpendicular to the longitudinal axis of laths.

3 Results

3.1 Microstructures

Figure 2 (a) through (f) show the microstructures of Ti624x (x=2, 4 and 6) alloys after two types of β annealing treatments, corresponding to normal (top row) and extra-coarse (bottom row) prior β grains. The microstructures exhibited here consisted of aligned α and β laths, which belong to typical fully lamellar microstructures. The microstructural size details after β annealing treatments have been summarized in Table 1.

For the relatively similar annealing temperature ($T_{\beta} + 20$ °C) or similar prior β grain size (~ 2090 μm), it was clear that both the α/β colony and α lath size decreased as Mo content was increased from Ti6242 to Ti6246. Besides the obvious variation of microstructural size with alloy types, annealing temperature obviously also affected the microstructural parameters, e.g., the higher temperature treatment produced larger prior β grains and α/β colonies but relatively thinner α laths. This applied to all three Ti624x alloys but to different degrees. That a coarser prior β grain would lead to thinner α lamellae is not intuitive although it has been reported in earlier research [50, 51].

Fully lamellar Ti microstructures with a β grain size of ~ 600 μm in titanium alloys are widely reported in the open literature and are usually obtained by annealing at a temperature within 30-50 °C above T_{β} [28], compared with $T_{\beta} + 20$ °C heat-treated microstructure in the present study. However, it is unusual to have a microstructure with an average prior β grain size as large as ~ 2090 μm obtained by beta heat treatment. Such a grain size was comparable to the microstructures of some cast titanium components (~ 1.5 mm) [52, 53], but still much smaller than that of the production-scale cast ingots (typically > 10 mm) [54]. So the lamellar microstructure with extra-coarse β grains obtained in the current study was a good model microstructure, to some extent, for simulating the FCG behavior in the large casting microstructure. In addition, FCG data corresponding to the extra-coarse β grain microstructure could directly be compared with the ‘normal’ β grain microstructure, which is the standard microstructure, to bring out the effect of microstructural parameters on FCG behavior.

3.2 Fatigue Crack Growth Rates

The experimental data for all the conditions were plotted on the log-log diagrams in the form of FCG rate (da/dN) versus ΔK , as shown in Figure 3, where the initial value of ΔK is about 10 MPa $\sqrt{m}$. The FCG curves in this figure were grouped by alloy types so as to exhibit the effect of stress ratio and microstructures. The FCG behavior investigated here was in Paris regime, where the measured data points could be described and fitted by the Paris equation [36],

$$da/dN = C(\Delta K)^m \quad (1)$$

where a is the crack length, N is the number of loading cycles, C and m are parameters resulting from the best-fitting lines. The specific values of C and m , for all the conditions investigated in the present study, were obtained as listed in Table 2. It should be noted that there was a transition phenomenon in FCG rate curves for Ti6242 and Ti6244 alloys with extra-coarse β grains tested under $R = 0.1$, as indicated by Letter ‘T’ in Figure 3(a) and (b), where two stages were adopted to do the fitting.

Figure 3 showed that both the stress ratio and microstructure affected the final FCG rate response. Their effects could be quantitatively compared to some extent by choosing the FCG rate data at a specific ΔK value and calculating the changes in FCG rate, such as $(FCG_{0.7}-FCG_{0.1})/FCG_{0.1}$ and $(FCG_{\text{extra}}-FCG_{\text{normal}})/FCG_{\text{normal}}$. The comparison, conducted on Ti6242 alloy and presented in Table 3, showed that increasing the microstructural unit size (mainly referring to α/β colony size) would reduce the FCG rate, more significantly at $R=0.1$. And increasing the stress ratio would enhance the FCG rate, more significantly for the coarser microstructure. The changes in FCG rate caused by increasing stress ratio were much larger compared to those due to microstructure variation.

3.3 Fatigue Crack Growth Paths

In order to clarify the crack growth micro-mechanisms, the FCG paths were examined by OM technique along the middle plane parallel to the CT sample broad surfaces. Figures 4 and 5 show the low magnification of FCG path profiles of Ti624x alloys tested under $R = 0.1$ and 0.7, respectively, including both normal and extra-coarse β grain microstructures. The FCG paths could be described and compared from two aspects, namely crack deflection range and bifurcation degree. The deflection range of

a crack path was termed as d which was indicated in Figure 4(a). The crack deflection could be compared quantitatively using the measured deflection range data as summarized in Table 4.

Three characteristics could be obtained after a careful observation and comparison of these path profiles: Firstly, while tested under $R = 0.1$, the extra-coarse β grain microstructures, as shown in Figure 4(b), (d) and (f), corresponded to larger crack deflection than normal β grain microstructures, as shown in Figure 4(a), (c) and (e). And an obvious decreasing of the difference between the two types of microstructures in crack deflection range could be found from Ti6242 to Ti6246. Secondly, while tested under $R = 0.7$, the extra-coarse β grain microstructures, *e.g.*, Figure 5(b), (d) and (f), have similar crack deflection to normal β grain microstructures, *i.e.*, Figure 5(a), (c) and (e), and very small differences existed in different alloys. Overall, the deflection range of crack paths tested under $R = 0.7$ was smaller than those tested under $R = 0.1$. Thirdly, the crack paths of $R = 0.7$ presented obvious bifurcation, as illustrated in Figure 5(a), (d) and (e), where in some locations the main crack propagated forth along two paths. Additionally, the crack paths under a high stress ratio showed relatively larger local roughness and gap between both crack surfaces than a low stress ratio.

The crack path images clearly showed that the FCG behavior was strongly influenced by the microstructures and stress ratio. Both decreasing the microstructural size and increasing the stress ratio would lead to a flatter crack path and smaller resistance to crack growth. For similar average prior β grain size in Ti624x alloys, as shown in Figure 4(b), (d) and (f), the different crack deflection illustrated that the fracture surface roughness was not controlled by β grains, but by perhaps α/β colonies or α laths. The FCG behavior reflected in growth paths was consistent with the FCG rates in Figure 3.

4 Discussion

4.1 Effects of Crystallographic Orientation and Microstructure

In the early stage of FCG in titanium alloys, *i.e.* near-threshold region, the fatigue cracking has been demonstrated to be sensitive to microstructure details and the crack

usually propagates along planar slip bands in α/β colonies, which resulted in the significant RICC effect [10, 11, 39, 55, 56]. In the case of fully lamellar microstructures, particular emphasis should be placed on the role of the α/β colony size in affecting the FCG behavior, wherein the crystallographically aligned individual α laths within a colony appear to behave like a single crystal. So the large colonies usually lead to long and continuous slips or cracks, which is usually accompanied by more crack deflection, higher fracture surface roughness and consequently slower crack growth [9, 57], such as the case of Ti6242 with extra-coarse prior β grain microstructure tested under $R = 0.1$ in the current study. Therefore, both the α/β colony size and its orientation are important factors in influencing the FCG behavior in the early stage (at low ΔK). However, for the later stage with larger crack lengths, FCG behavior was less sensitive to the microstructure, where fatigue crack occurs in non-preferably oriented α/β colonies and travels across the colonies in striated growth [11], thus with a decreased RICC.

A small representative section of the early stage region of the FCG path was chosen to analyze the interaction of α/β colony orientation and crack path. Figure 6 through 8, corresponding to Ti6242, Ti6244 and Ti6246, respectively, showed the specific characterization results including all Euler map (a), OM image (b) and schematic 3D hexagons (c). The labeled schematic 3D hexagons, as shown in Figure 6(c) through 8(c), denoted the crystallographic orientation of the corresponding labeled α/β colonies for better visualization of the orientation relationship with the crack path.

Close inspection of the colony orientation and crack path in Figure 6 through 8 showed that the fatigue cracks mainly propagated along the basal planes of α lamellae and sometime along prismatic planes (only illustrated as Colony 3 in Figure 6, Colony 1 in Figure 7 and Colony 5 in Figure 8). Here transverse-to-lamella cracks are the most common cases, and the normal of the basal plane is close to a 45° angle to loading direction (and as such consistent with ‘invariant line’ preferential platelet growth direction), although parallel-to-lamella is also an important failure mode according to the statistics of the angles between the crack path and the α lamella long axis direction, which partly depends on the loading frequency [10, 11, 18].

Throughout the crack paths in Figure 6 through 8, the cracks in the early stage propagated forward in accordance with a zigzag path: The cracks grow preferentially on basal planes resulting in the crack path tortuosity from colony to colony. As indicated by the microstructural features in Table 1, it was shown that the α/β colony size obtained for both heat treatments reduced as Mo content rose in the alloys. This has been clearly demonstrated by the all Euler maps in Figure 6(a) through 8(a). Smaller α/β colonies mean shorter slip length, easier diversion and thus lower roughness. So both the α/β colony orientation and colony size affected the path deflection in the early stage of FCG. When the crack continued to propagate, it corresponded to a higher crack-driving force and higher FCG rates. The crack propagation gradually switched to the striation propagation [11], where the FCG occurred in non-preferably oriented colonies and travelled straight through α/β colonies nearly without deviating. In higher ΔK regime, the microstructure-insensitive cracking dominated over the crystallographic cracking, which led to a decreased fracture surface roughness, as typically shown in Figure 4 (b) and (d).

Regarding the variation of microstructure unit size in Ti624x (x=2, 4 and 6) alloys, two aspects should be considered to account for the phenomenon observed in the current study. Firstly, as discussed in our previous study [46], with the increase of Mo content from Ti6242 to Ti6246, the ‘undercooling’ due to reduced T_β temperature gradually increases, encouraging the nucleation of α laths. The increasing Mo content also hinders diffusion and slows down α lath coarsening. As a result, the size of the final α laths and α/β colonies decrease steadily from Ti6242 to Ti6246. Secondly, the phenomenon that thinner α laths appeared in larger prior β grains for each alloy was usually explained as due to a larger β grain having a lower $\beta \rightarrow \alpha$ transformation temperature and thus reduced diffusivity of the alloying elements [50, 51], which may itself be due to segregation of α stabilisers to the grain boundary. A more specific hypothesis is that larger β grains, obtained through heat treatment at higher solution temperature or for longer solution time, correspond to fewer residual defects and thus the diffusion of alloying elements are decelerated [47, 58]. The specific mechanism is still unclear and needs further investigation.

4.2 Effect of K_{max}

The microstructure (primarily referring to the α/β colony size) was an important aspect in determining the FCG behavior, as we discussed in Section 4.1. However, a comparison of the FCG paths of two types of microstructures tested under $R = 0.7$ in Figure 5 suggested that the difference of FCG paths due to the microstructure was diminished, which was also confirmed by the FCG rate curves of $R = 0.7$ in Figure 3.

Stress ratio determines the relationship between K_{max} and ΔK . For a given ΔK , stress ratio effect is equivalent to K_{max} effect to a large extent. K_{max} has been extensively considered in fatigue life prediction as a factor in addition to ΔK [12, 16, 32, 33]. As pointed out by Sadananda et al. [13], the contribution of K_{max} dependent process to FCG should be considered because considering only the ΔK contribution would not give a reliable life prediction. However, its specific influence on the failure mechanisms has been rarely analyzed until now. Usually it was thought that the stress-field in front of the crack tip causes the dislocation slip and fatigue crack formation in preferably orientated α/β colonies, where K_{max} and ΔK play different roles. Sadananda et al. thought that K_{max} associated with the monotonic plastic zone (MPZ) essentially governs the breaking of the bonds contributing to crack growth, while ΔK corresponding to the cyclic plastic zone (CPZ) controls the extent of cyclic damage required ahead of the crack tip [13].

As we stated in section 3.3, a higher stress ratio was associated with a flatter crack path, relatively larger local roughness and gap between both crack surfaces, which was much more obvious when we compared the FCG paths of extra-coarse microstructures tested under $R = 0.1$ and 0.7 . In the current study, the increase of ΔK is identical for different stress ratios as cracks extended forward, where only K_{max} having different values could account for the obvious difference in the FCG paths. So in this section, the K_{max} effect on slip cracking behavior would be clarified from two aspects including plastic zone size and stress levels ahead of the crack tip.

Firstly the plastic zone size under different stress ratios could be determined by the analytical estimation (suitable for the plain strain condition) [59], as follows:

$$r_p^M = (K_{max}/\sigma_y)^2 / (3\pi) \quad (2)$$

$$r_p^C = (\Delta K / 2\sigma_y)^2 / (3\pi) \quad (3)$$

where r_p^M and r_p^C correspond to the size of MPZ and CPZ, respectively. ΔK and K_{max} are derived according to LEFM. σ_y is the yield strength of the material.

The peak applied loads are 5 and 13 kN corresponding to $R = 0.1$ and 0.7 , respectively. According to equation (2) and (3), the schematic comparison of CPZ and MPZ size is illustrated in Figure 9(a), which shows the proportional relationship as follows: $CPZ_{0.1/0.7} : MPZ_{0.1} : MPZ_{0.7} = 81 : 400 : 3600$. The specific estimated value of plastic zone size, taking Ti6244 alloy with extra-coarse prior β grain microstructure as an example, has been given in Figure 9(b), which shows the size evolved with ΔK . The initial values of $CPZ_{0.1/0.7}$, $MPZ_{0.1}$ and $MPZ_{0.7}$ size (at ΔK of about $10.5 \text{ MPa}\sqrt{\text{m}}$) are about 0.004, 0.020 and 0.178 mm, respectively. The MPZ size under $R = 0.7$ was much larger than the other plastic zone values, *i.e.* $CPZ_{0.1/0.7}$ and $MPZ_{0.1}$. So, under higher stress ratio, the dislocation slip behavior happens in a much larger area ahead of the crack tip. Suppose the fatigue crack under $R = 0.7$ propagates through the $MPZ_{0.7}$ (as large as $\sim 0.178 \text{ mm}$) at the rate of initial stage ($\sim 1.1 \times 10^{-5} \text{ mm/cycle}$), it needs about 1.6×10^4 cycles of life. But for the crack propagation under $R = 0.1$, it needs fewer cycles ($\sim 4.2 \times 10^3$ cycles) to grow through the $MPZ_{0.1}$ (as large as $\sim 0.02 \text{ mm}$) at the rate of its initial stage ($\sim 4.8 \times 10^{-6} \text{ mm/cycle}$). So in the preferably orientated α/β colonies ahead of the fatigue crack, although both basal and prismatic slips can operate and be developed into slip bands, the longer fatigue life (under higher K_{max}) means more sufficient development and easier crack forming along multiple slip planes.

The other aspect of the K_{max} effect on slip cracking behavior may be related to the stress and strain condition ahead of the crack tip which governs crack progression. With the hypothesis of LEFM, the stresses near the crack tip could be calculated according to the following equation [60]:

$$\sigma_{r\theta} = \frac{K}{\sqrt{2\pi r}} f(\theta) \quad (4)$$

where r and θ are polar coordinates ahead of the crack tip and K is the stress intensity

factor, $f(\theta)$ is the general trigonometric function.

For CT samples in this study, K could be further calculated from Equations (5a) and (5b) per ASTM E-399-90 standard [61].

$$K = (P/BW^{1/2}) \cdot f(a/W) \quad (5a)$$

$$f(a/W) = \frac{(2+a/W)(0.886+4.64a/W-13.32a^2/W^2+14.72a^3/W^3-5.6a^4/W^4)}{(1-a/W)^{3/2}} \quad (5b)$$

where P is the load, B and W are the thickness and width of the specimen and a is the crack length.

So, according to the above analysis, the stress levels ahead of the crack tip under different stress ratios were proportional to the load for the same alloy and specimen geometries. Specifically the peak stress under $R = 0.1$ and 0.7 was $\sigma_{0.1} : \sigma_{0.7} = 5 : 13$. However, the peak stress was limited by the flow stress, and given the typically low work hardening rates of titanium alloys, it would not be very much higher at $R = 0.7$. But the amount of accumulated plastic strain ahead of the crack tip under $R = 0.7$ should be larger than under $R = 0.1$. In this case, the higher K_{max} corresponding to a higher maximum stress and larger strain would more easily initiate cracks within both basal and prism slip bands, as shown in Figure 10(b) and (d), where the crack could propagate forward with heavy bifurcation even though in a single α/β colony. However, under $R = 0.1$ the planar crack occurred along only one slip direction, as shown in Figure 10(a) and (c). So during the FCG process under $R = 0.7$, even though in large colonies the crack may bifurcate easily and grow alternately along two directions, resulting in the small deflection in the crack growth path, as shown in Figure 5(b), (d) and (f). While the crack propagated forward like the situation in Figure 10(a) and (c), it would make the crack path locally very straight, and finally appeared with a large deflection macroscopically, as shown in Figure 4(b) and (d). Therefore, besides its role in determining the crack closure effect as usually reported, the stress ratio also affected the cracking behavior ahead of the crack tip (referring mainly to the occurrence of bifurcation).

4.3 Effect of crack closure

As we discussed in section 4.1 and 4.2, both the microstructure and K_{max} have effects

on FCG paths. The smaller α/β colonies corresponded to shorter cleavage crack length and smaller path roughness, and a higher K_{max} was associated with significant bifurcation and also reduced path roughness. Both cases thus resulted in smaller RICC effect and increased the FCG rates. In addition to the bifurcation, the high K_{max} corresponding to high stress ratio also reduced the RICC effect through increasing the opening displacement between crack surfaces to reduce their contact [62]. As we know, the crack closure effect arises from the smaller crack tip opening displacements which enhance the propensity for the interference of asperities along crack flanks and effectively shield the crack tip from the far field stress intensity range. Such an effect can be enhanced by coarsening the microstructural unit size.

In the early FCG stage of the extra-coarse microstructures tested under $R = 0.1$, a cleavage propagation mode resulted in obvious tortuous crack growth and a large degree of crack deflection and thus was accompanied by high levels of crack closure, as shown in Figure 4(b) and (d). Under this circumstance, a transition point phenomenon occurred in the da/dN vs. ΔK curves, as illustrated in Figure 3(a) and (b). Smaller α/β colony sizes corresponded to reduced critical transitions Δk . In titanium alloys, the critical transition ΔK is associated with the crack tip plasticity zone exceeded the α/β colony dimension [39, 63]. Because an increased probability of symmetric slip process appeared due to different orientation in the surrounding microstructural units, the propagation changed from a cleavage to a striation mode [30]. However, taking Ti6244 in this study as an example, we found that the CPZ and $MPZ_{0.1}$ size at the transition point of $\Delta K = 20.2 \text{ MPa}\sqrt{\text{m}}$ were ~ 15.9 and $78.7 \text{ }\mu\text{m}$, respectively. Both of the plasticity zone sizes were much smaller than the α/β colony size ($\sim 382.6 \text{ }\mu\text{m}$). The reason may be that the plastic zone size calculations do not reflect the true colony behavior in the appropriate loading direction as they are derived from bulk properties. However, it could be confirmed that the transition phenomenon actually occurred after the cleavage propagation with an obvious tortuous path, as defined by the labeled ‘T’ in Figure 4(b) and (d). Thus, the low FCG rates near the transition point were primarily caused by the large crack closure effect.

In order to compare the damage tolerance properties of Ti624x alloys, FCG rate

curves were grouped by the heat treatment and loading conditions, which were shown in Figure 11 and 12, respectively. The results showed that the damage tolerance properties gradually decreased from Ti6242 to Ti6246.

A careful comparison of the FCG rate curves in different alloys showed that the main difference appeared in the early stage of Paris-region because the FCG rate is much more sensitive to the microstructure. For Ti624x alloys tested under the same loading conditions, the difference of FCG behavior depended on their α/β colony size, where the prior β grain effect could be excluded from the results in Figure 12.

The crack closure effect was essentially absent if the tests were conducted under high stress ratio [64, 65]. The difference of FCG rates under $R = 0.7$ in Ti624x alloys, as shown in Figure 11(b) and 12(b), was still caused by the variable α/β colony sizes because more tortuous paths (as shown in Table 4) consumed more energy in the absence of crack closure, and this was another factor reducing the FCG rates.

For normal and extra-coarse prior β grain microstructures, α laths have similar decreasing size with the increasing of Mo content; however, comparing the FCG rate curves in Figure 11(a) and 12(a), Figure 11(b) and 12(b), we see a larger difference of FCG rates in extra-coarse microstructures than the normal ones for the early stage of Paris-region, which indirectly demonstrated again that α/β colony played a more decisive role than α lath in long crack FCG behavior of titanium alloys.

5 Conclusions

The effect of microstructures and stress ratios on long crack FCG behavior of titanium alloys has been investigated using Ti624x ($x = 2, 4$ and 6) alloys in this study. Through the heat treatments, two types of lamellar microstructures with the normal and extra-coarse prior β grains have been prepared. FCG tests under $R = 0.1$ and 0.7 were conducted to analyze the effects of the K_{max} and crack closure from the FCG rate curves, crack paths, and the plastic zone ahead of the crack tip. The conclusions reached in this study are as follows.

1. With similar heat treatments, a large variation of microstructural parameters appeared in Ti624x ($x = 2, 4$ and 6) alloys. The size of prior β grain, α/β colony and α laths all decreased with the increasing of Mo content. Even if with similar

prior β grain size, the increase of the content of β stabilizing element Mo decreased the tendency to transform the microstructure into large α/β colonies. This was because the increase of Mo content in Ti624x alloys would encourage the nucleation of α laths at lower temperature, hinder the diffusion and slow down their coarsening. α lath size was also influenced by β grain size in addition to usual factors, such as the cooling rate as usually reported.

2. The damage tolerance properties (about FCG) from Ti6242 to Ti6246 alloys were degraded with increasing Mo content. This was driven by the variation of α/β colony size. The property discrepancy was more obvious for the early stage of Paris-region, where FCG behavior was sensitive to microstructure and usually resulted in the cleavage propagation and zig-zag paths. Larger α/β colonies would lead to higher crack deflection and slower FCG rates.
3. In addition to microstructure effect, stress ratios also affected the FCG behavior where a higher stress ratio corresponded to lower crack deflection and higher FCG rates. This phenomenon could be attributed to two drivers: one is that increasing the stress ratio would reduce the contact of fracture surfaces and increase the ΔK_{eff} , the other is that a higher stress ratio in this study corresponded to a higher K_{max} and resulted in heavier crack bifurcation and propagation alternately along two directions, which finally led to the relatively smooth crack growth path.
4. A transition phenomenon on the FCG rate curves occurred in the extra-coarse microstructures of Ti6242 and Ti6244 alloys which was tested under $R = 0.1$. This was because in the early stage of Paris-region, long and straight cleavage propagation under low stress ratio caused strong RICC effect and obviously reduced the FCG rates.

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References

- [1] Boyer RR. An overview on the use of titanium in the aerospace industry. *Mater Sci Eng A* 1996; 213: 103-14.
- [2] Williams JC, Starke EA. Progress in structural materials for aerospace systems. *Acta Mater* 2003; 51: 5775-99.
- [3] Wood HA, Engle RM. *USAF Damage Tolerant Design Handbook: Guidelines for the Analysis and Design of Damage Tolerant Aircraft*. United States Air Force, 1979.
- [4] EASA CM-PIFS-007 Issue: 01, Engine Critical Parts-Damage Tolerance Assessment Manufacturing and Surface Induced Anomalies. European Aviation Safety Agency, 2013.
- [5] Larsen J, Worth B, Annis C, Haake F. An assessment of the role of near-threshold crack growth in high-cycle-fatigue life prediction of aerospace titanium alloys under turbine engine spectra. *Int J Fract* 1989; 80: 237-55.
- [6] B.E P. Fatigue crack growth behaviour of two contrasting titanium alloys. *Int J Fatigue* 1995; 17: 221-7.
- [7] Saxena VK, Radhakrishnan VM. Effect of phase morphology on fatigue crack growth behavior of alpha-beta titanium alloy - A crack closure rationale. *Metall Mater Trans A* 1998; 29: 245-61.
- [8] Jha SK, Ravichandran KS. Effect of mean stress (stress ratio) and aging on fatigue-crack growth in a metastable beta titanium alloy, Ti-10V-2Fe-3Al. *Metall Mater Trans A* 2000; 31: 703-14.
- [9] Jin O, Mall S. Effects of microstructure on short crack growth behavior of Ti-6Al-2Sn-4Zr-2Mo-0.1Si alloy. *Mater Sci Eng A* 2003; 359: 356-67.
- [10] Sansoz F, Ghonem H. Fatigue crack growth mechanisms in Ti6242 lamellar microstructures: Influence of loading frequency and temperature. *Metall Mater Trans A* 2003; 34A: 2565-77.
- [11] Sansoz F, Ghonem H. Effects of loading frequency on fatigue crack growth mechanisms in α/β Ti microstructure with large colony size. *Mater Sci Eng A* 2003; 356: 81-92.
- [12] Ding J, Hall R, Byrne J. Effects of stress ratio and temperature on fatigue crack growth in a Ti-6Al-4V alloy. *Int J Fatigue* 2005; 27: 1551-8.
- [13] Sadananda K, Vasudevan AK. Fatigue crack growth behavior of titanium alloys. *Int J Fatigue* 2005; 27: 1255-66.
- [14] Benedetti M, Heidemann J, Peters JO, Lütjering G. Influence of sharp microstructural gradients on the fatigue crack growth resistance of $\alpha+\beta$ and near- α titanium alloys. *Fatigue Fract Eng Mater Struct* 2005; 28: 909-22.
- [15] Adib AML, Baptista CARP. An exponential equation of fatigue crack growth in titanium. *Mater Sci Eng A* 2007; 452-453: 321-5.
- [16] Noroozi AH, Glinka G, Lambert S. A study of the stress ratio effects on fatigue crack growth using the unified two-parameter fatigue crack growth driving force. *Int J Fatigue* 2007; 29: 1616-33.
- [17] Korsunsky AM, Song X, Belnoue J, Jun T, Hofmann F, De Matos PFP, Nowell D, Dini D, Aparicio-Blanco O, Walsh MJ. Crack tip deformation fields and fatigue crack growth rates in Ti-6Al-4V. *Int J Fatigue* 2009; 31: 1771-9.
- [18] Ghonem H. Microstructure and fatigue crack growth mechanisms in high temperature titanium alloys. *Int J Fatigue* 2010; 32: 1448-60.
- [19] Walker KF, Newman JC. Improved test method and analytical modelling for fatigue crack growth in coarse-grain titanium alloy with rough fatigue surfaces. *Fatigue Fract Eng Mater Struct* 2014; 37: 659-70.
- [20] Ghonem H, Foerch R. Frequency effects on fatigue crack growth behavior in a near alpha titanium

alloy. Mater Sci Eng A 1991; 138: 69-81.

[21] Oberwinkler B. Modeling the fatigue crack growth behavior of Ti-6Al-4V by considering grain size and stress ratio. Mater Sci Eng A 2011; 528: 5983-92.

[22] Shaniavski AA, Losev AI. The effects of loading waveform and microstructure on the fatigue response of titanium aero-engine compressor disk alloys. Fatigue Fract Eng Mater Struct 2003; 26: 329-42.

[23] Belnoue JP, Jun TS, Hofmann F, Abbey B, Korsunsky AM. Evaluation of the overload effect on fatigue crack growth with the help of synchrotron XRD strain mapping. Eng Fract Mech 2010; 77: 3216-26.

[24] Bache MR, Evans WJ, McElhone M. The effects of environment and internal oxygen on fatigue crack propagation in Ti-6Al-4V. Mater Sci Eng A 1997; 234: 918-22.

[25] Petit J, Sarrazin-Baudoux C. Some critical aspects of low rate fatigue crack propagation in metallic materials. Int J Fatigue 2010; 32: 962-70.

[26] Gray GT, Luetjering G, Williams JC. The influence of oxygen on the structure, fracture, and fatigue crack propagation behavior of Ti-8.6 Wt Pct Al. Metall Trans A 1990; 21: 95-105.

[27] Leyens C, Peters M. Titanium and Titanium Alloys: Fundamentals and Applications. Wiley; 2003.

[28] Lütjering G, Williams JC. Titanium. Springer; 2007.

[29] Biroasca S, Buffiere JY, Karadge M, Preuss M. 3-D observations of short fatigue crack interaction with lamellar and duplex microstructures in a two-phase titanium alloy. Acta Mater 2011; 59: 1510-22.

[30] Dubey S, Soboyejo ABO, Soboyejo WO. An investigation of the effects of stress ratio and crack closure on the micromechanisms of fatigue crack growth in Ti-6Al-4V. Acta Mater 1997; 45: 2777-87.

[31] Bania PJ, Eylon D. Fatigue crack propagation of titanium alloys under dwell-time conditions. Metall Trans A 1978; 9: 847-55.

[32] Kujawski D. A new $(\Delta K + K_{\max})^{0.5}$ driving force parameter for crack growth in aluminum alloys. Int J Fatigue 2001; 23: 733-40.

[33] Stoychev S, Kujawski D. Analysis of crack propagation using ΔK and $K_{\max}$. Int J Fatigue 2005; 27: 1425-31.

[34] Xiong Y, Katsuta J, Kawano K, Sakiyama T. Examination of fatigue crack driving force parameter. Fatigue Fract Eng Mater Struct 2008; 31: 754-65.

[35] Gudladt H-J, Steinbock J. Influence of $K_{\max}$ and R on Fatigue Crack Growth—A 3D-Model. Adv Eng Mater 2010; 12: 283-7.

[36] Paris P, Erdogan F. A critical analysis of crack propagation laws. Trans Am Soc Mech Eng J Basic Eng 1963; 528-34.

[37] Ravichandran KS. Near threshold fatigue crack growth behavior of a titanium alloy: Ti-6Al-4V. Acta Metall Mater 1991; 39: 401-10.

[38] Wanhill R, Barter S. Fatigue of Beta Processed and Beta Heat-treated Titanium Alloys. Springer; 2011.

[39] Yoder GR, Cooley LA, Crooker TW. Observations on microstructurally sensitive fatigue crack growth in a Widmanstätten Ti-6Al-4V alloy. Metall Trans A 1977; 8: 1737-43.

[40] Ravichandran KS. Fatigue crack closure as influenced by microstructure in Ti-6Al-4V. Scripta Metall Mater 1990; 24: 1559-63.

[41] Powell BE. Fatigue crack growth behaviour of two contrasting titanium alloys. Int J Fatigue 1995; 17: 221-7.

[42] Bathias C, El Alami K, Wu TY. Influence of mean stress on Ti6Al4V fatigue crack growth at very high

frequency. Eng Fract Mech 1997; 56: 255-64.

[43] Campbell JP, Ritchie RO. Mixed-mode fatigue-crack growth thresholds in Ti-6Al-4V at high frequency. Scripta Mater 1999; 41: 1067-71.

[44] Peralta P, Villarreal T, Atodaria I, Chattopadhyay A. Mechanical length scales and their link to fatigue crack growth kinetics in beta-annealed Ti-6Al-4V. Scripta Mater 2012; 66: 13-6.

[45] Kobryn PA, Semiatin SL. Microstructure and texture evolution during solidification processing of Ti-6Al-4V. J Mater Process Technol 2003; 135: 330-9.

[46] Qiu JK, Ma YJ, Lei JF, Liu YY, Huang AJ, Rugg D, Yang R. A comparative study on dwell fatigue of Ti-6Al-2Sn-4Zr-xMo (x = 2 to 6) alloys on a microstructure-normalized basis. Metall Mater Trans A 2014; 45: 6075-87.

[47] Qiu JK, Ma YJ, Ji HB, Lei JF, Liu YY, Yang R. Effect of Mo content on grain growth behaviour and mechanical properties of titanium alloys. Chin J Nonferrous Met 2013; 23: s153-s8.

[48] ASTM E647-08, Standard Test Method for Measurement of Fatigue Crack Growth Rates. ASTM International, 2008.

[49] ASTM E112-96, Standard Test Methods for Determining Average Grain Size. ASTM International, 1996.

[50] Gil FJ, Ginebra MP, Manero JM, Planell JA. Formation of alpha-Widmanstatten structure: effects of grain size and cooling rate on the Widmanstatten morphologies and on the mechanical properties in Ti6Al4V alloy. J Alloys Compd 2001; 329: 142-52.

[51] Lenain A, Clement N, Veron M, Jacques PJ. Characterization of the alpha phase nucleation in a two-phase metastable beta titanium alloy. J Mater Eng Perform 2005; 14: 722-7.

[52] Pilchak AL, Juhas MC, Williams JC. Microstructural changes due to friction stir processing of investment-cast Ti-6Al-4V. Metall Mater Trans A 2007; 38A: 401-8.

[53] Feng X, Peng S, Ma YJ, Lei JF, Wu XH, Cui YY, Yang R. Influences of hot isostatic pressing on fatigue crack propagation of TC4 titanium alloy thin wall casting. Chin J Nonferrous Met 2013; 23: s462-s6.

[54] Sargent GA, Kinsel KT, Pilchak AL, Salem AA, Semiatin SL. Variant selection during cooling after beta annealing of Ti-6Al-4V ingot material. Metall Mater Trans A 2012; 43A: 3570-85.

[55] Tanaka K, Mura T. Fatigue crack-growth along planar slip bands. Acta Metall 1984; 32: 1731-40.

[56] Ravichandran KS, Dwarakadasa ES. Theoretical modelling of the effects of grain size on the threshold for fatigue crack growth. Acta Metall Mater 1991; 39: 1343-57.

[57] Wang S-H, Müller C. Fracture surface roughness and roughness-induced fatigue crack closure in Ti-2.5 wt% Cu. Mater Sci Eng A 1998; 255: 7-15.

[58] Ivasishin OM, Markovsky PE, Semiatin SL, Ward CH. Aging response of coarse- and fine-grained β titanium alloys. Mater Sci Eng A 2005; 405: 296-305.

[59] Suresh S. Fatigue of Materials. Cambridge University Press; 1998.

[60] Meyers MA, Chawla KK. Mechanical Behavior of Materials: Second Edition. Cambridge University Press; 2009.

[61] ASTM E399-90, Standard Test Method for Plane-Strain Fracture Toughness of Metallic Materials. ASTM International, 1997.

[62] Suresh S, Ritchie RO. A geometric model for fatigue crack closure induced by fracture surface roughness. Metall Trans A 1982; 13: 1627-31.

[63] Yoder GR, Cooley LA, Crooker TW. Fatigue crack propagation resistance of beta-annealed Ti-6Al-4V alloys of differing interstitial oxygen contents. Metall Trans A 1978; 9: 1413-20.

[64] Chan K. Variability of large-crack fatigue-crack-growth thresholds in structural alloys. Metall Mater

Trans A 2004; 35: 3721-35.

[65] Yamada Y, Newman Jr JC. Crack closure under high load ratio and test conditions. Procedia Engineering 2010; 2: 71-82.

Table 1. Details of β annealing treatments and the respective microstructural sizes for Ti624x alloys investigated.

Alloy	β annealing (°C)	β grain size (μm)		Colony size (μm)		α lath size (μm)	
		Mean	S. D.*	Mean	S. D.**	Mean	S. D.**
Ti6242	T β +20	887.2	381.4	241.4	23.8	2.7	0.4
	1240	2082.6	1107.6	447.1	43.2	2.1	0.3
Ti6244	T β +20	644.3	303.0	178.2	16.3	1.8	0.2
	1300	2095.8	1080.8	382.6	71.7	1.2	0.2
Ti6246	T β +20	464.1	213.4	97.4	5.0	1.4	0.2
	1300	2095.1	1273.2	143.2	22.8	1.1	0.2

*Measured using the equivalent method, taking each grain as the individual.

**Measured using the intercept method, taking the average of each view field as the individual.

Table 2. The fitting parameters of Paris equation used to fit the measured FCG data.

Stress Ratio	Alloys	T _β +20°C		2090 μm	
		C [m/(Cyc. MPa√m)]	m (-)	C [m/(Cyc. MPa√m)]	m (-)
R=0.1	Ti6242	4.15E-08	4.25	2.65E-07	2.55
				6.46E-09	5.34
	Ti6244	6.13E-08	4.04	2.46E-06	0.67
				3.56E-08	4.23
				1.43E-07	3.53
R=0.7	Ti6242	1.54E-07	3.65	1.11E-07	3.90
	Ti6244	1.32E-07	3.92	1.81E-07	3.60
	Ti6246	1.23E-07	4.03	3.43E-07	3.15

Table 3. Comparison of the microstructure and stress ratio effect on the FCG rates.

The data were obtained at $\Delta K = 15 \text{ MPa}\sqrt{\text{m}}$ of Ti6242 alloy.

Microstructure effect		Stress ratio effect	
$R = 0.1$	$R = 0.7$	$T_{\beta} + 20 \text{ }^{\circ}\text{C}$	$\sim 2090 \text{ }\mu\text{m}$
-27.6%	-5.9%	254.7%	361.1%

Table 4. The deflection range of FCG paths measured with the method as defined in Figure 4(a).

Stress Ratio	Alloys	T+20°C (mm)	2090 μm (mm)
R=0.1	Ti6242	3.43	5.58
	Ti6244	1.34	2.25
	Ti6246	1.29	1.52
R=0.7	Ti6242	3.46	2.3
	Ti6244	1.24	1.93
	Ti6246	1.47	1.47

Figure 1. Compact-tension sample for FCG tests with the geometry indicated ($W = 40$ mm, $B = 10$ mm).

Figure 2. Backscatter electron images of Ti624x fully lamellar microstructures, which indicate the variation of α colonies and platelets with alloy types and heat treatments: (a) (d) Ti6242, (b) (e) Ti6244, (c) (f) Ti6246; (a)-(c) after solution treatment at $T_{\beta}+20^{\circ}\text{C}$ for the respective alloy, (d)-(f) after solution treatment at different temperatures to achieve the same average prior β grain size of about $2090\text{ }\mu\text{m}$ in the three alloys.

Figure 3. FCG rates (da/dN) as a function of applied stress intensity range (ΔK) for Ti624x alloys tested under $R = 0.1$ and 0.7 : (a) Ti6242, (b) Ti6244 and (c) Ti6246. The solid lines were the fitting results according to Paris equation. Letter 'T' labeled in (a) and (b) indicated the transition point in the FCG rate curve. (Color figure online).

Figure 4. FCG path profiles of Ti624x alloys with different β annealing treatments tested under $R = 0.1$: (a, b) Ti6242, (c, d) Ti6244, (e, f) Ti6246; (a, c, e) after solution treatment at $T_{\beta}+20^{\circ}\text{C}$ for the respective alloy, (b, d, f) after solution treatment at different temperatures to achieve the same average prior β grain size of about $2090\text{ }\mu\text{m}$ in the three alloys. The roughness deflection of growth path is termed as d , as indicated in (a). The arrows labeled with 'T' on (b) and (d) indicate the locations corresponding to the transition point in FCG rate curves shown in Figure 3. The FCG direction is from right to left.

Figure 5. FCG path profiles of Ti624x alloys with different β annealing treatments tested under $R = 0.7$: (a, b) Ti6242, (c, d) Ti6244, (e, f) Ti6246; (a, c, e) after solution treatment at $T_{\beta}+20^{\circ}\text{C}$ for the respective alloy, (b, d, f) after solution treatment at different temperatures to achieve the same average prior β grain size of about $2090\text{ }\mu\text{m}$ in the three alloys. The FCG direction is from right to left.

Figure 6. Crystallographic orientation characterization of α colonies near the crack path of Ti6242 alloy tested under $R = 0.1$ located at $\Delta K = 12.3 \text{ MPa}\sqrt{\text{m}}$ ~~the initial stage of Paris regime~~.

(a) EBSD-derived all-Euler map, (b) the corresponding OM image and (c) 3D crystal viewer for the corresponding labeled α colony.

Figure 7. Crystallographic orientation characterization of α colonies near the crack path of Ti6244 alloy tested under $R = 0.1$ located at $\Delta K = 11.7 \text{ MPa}\sqrt{\text{m}}$ ~~the initial stage of Paris regime~~.

(a) EBSD-derived all-Euler map, (b) the corresponding OM image and (c) 3D crystal viewer for the corresponding labeled α colony.

Figure 8. Crystallographic orientation characterization of α colonies near the crack path of Ti6246 alloy tested under $R = 0.1$ located at $\Delta K = 11.3 \text{ MPa}\sqrt{\text{m}}$ ~~the initial stage of Paris regime~~.

(a) EBSD-derived all-Euler map, (b) the corresponding OM image and (c) 3D crystal viewer for the corresponding labeled α colony.

(a)

(b)

Figure 9. (a) Schematic of CPZ and MPZ extent ahead of the crack tip under the same ΔK but different stress ratios, with the load parallel to y; (b) typical example of Ti6244 with extra-coarse prior β grains showing the estimated radius size of CPZ and MPZ evolving with ΔK .

Figure 10. SEM images showing the variable FCG cracking morphology in Ti624x alloys with extra-coarse prior β grains tested under different stress ratios, (a) Ti6242, $R = 0.1$; (b) Ti6242, $R = 0.7$; (c) Ti6244, $R = 0.1$ and (d) Ti6244, $R = 0.7$. The FCG direction is from left to right.

Figure 11. Comparison of FCG rates for Ti624x alloys annealed at $T_{\beta}+20^{\circ}\text{C}$ tested under: (a) $R = 0.1$ and (b) $R = 0.7$.

Figure 12. Comparison of FCG rates for Ti624x alloys with the same average prior β grain size of about $2090 \mu\text{m}$ tested under: (a) $R = 0.1$ and (b) $R = 0.7$.