On the microstructure evolution during isothermal low cycle fatigue of β-annealed Ti-6242S titanium alloy: Internal damage mechanism, substructure development and early globularization

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Abstract

The present work deals with the room and high temperature microstructure evolutions of a near alpha titanium alloy (Ti-6242S) during isothermal low cycle fatigue. The initial fully lamellar microstructure has been purposefully annealed to form α/β interface phase layers. The shearing and displacement of the β layers is identified as the main governing internal damage mechanism. This is phenomenal owning to the very low imposed strain; and is justified considering: (i) the large colony size and thin thickness of the β layers within the initial lamellar microstructure, (ii) the presence of the interface phase, and (iii) the intensified substructure development within the α and β phases. The shearing and displacement of the β layers lead to the segmentation and this is followed by initial stage of dynamic globularization even at room temperature condition. The globularization kinetic is intensified at higher temperature cyclic loading. In this regard, the sub-boundary induced boundary splitting/grooving is characterized as an involved mechanism. Interestingly, the early dynamic globularization of the α lamellae is also traced through the high temperature fatigued microstructures. The β phase penetration along the α phase sub-boundary is realized as the main globularization mechanisms. The interface phase layer stimulates the possibility of globularization via providing the high-energy non-coherent interfaces in structure. In addition, the deformation twinning within the interface phase, so called as "twinning induced micro-plasticity effect", appears to have a significant effect on the fatigue damage behavior via compensation of incompatibility between α and β phases and deformation twin formation.

Previous article in issue Next article in issue

Keywords

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1. Introduction

In the era of fourth industrial revolution, the availability of materials which can tolerate the thermal and mechanical cyclic stresses during service conditions appears to be an outstanding privilege. In this respect, any severity in service condition may be accompanied by a higher plastic deformation and damage accumulation, specifically at elevated temperatures, and this in turn may lead to a total lifetime of only a few thousand large strain cycles [1], [2]. Accordingly, many researchers have attempted to assess the isothermal strain-controlled fatigue behavior of metallic materials, in particular at elevated temperatures [2], [3], [4], [5], [6], [7], [8]. In the case of titanium alloys, the isothermal fatigue properties are realized to be highly sensitive to the involved microstructural features including prior β grain size, α + β colony size, and the thickness of α lamellae [9], [10], [11], [12]. It has been shown that the fully lamellar microstructure with a considerable fraction of α/β interface, possess a higher resistance to fatigue crack propagation in comparison to an equiaxed one where the applied cyclic stress level is high and the low cycle fatigue is dominant [13]. The latter clearly expresses the significance of interface characteristics on the fatigue behavior of lamellar microstructures. In this regard, the presence of α/β interface phase as an intermediate phase, which generally exists in fully lamellar microstructure that is cooled by relatively low cooling rate [14], [15], [16] would definitely introduce a significant effect on the room and high temperature fatigue properties. The latter has been mainly overlooked in previous works, therefore a set of systematic and comprehensive researches are highly necessitated. Zhao et al. [17] has previously found that the interface phase has a significant effect on the tensile properties of Ti-64 alloy, hinders the dislocation movement thereby decreasing the effective slip length. Additionally, it is believed that the deformation twinning may activate within the α/β interface phase during tensile deformation and acts similar to grain boundaries against dislocation movement [17].

Another worthy point which should be considered here is the presence of various deformation bands and sub-boundaries in the microstructures which are monotonically deformed at high temperatures [18], [19], [20], [21]. These would affect the evolution of lamellae phases, trigger the occurrence of lamellae phases shearing, lamellae grooving [22], [23] and subsequent dynamic globularization. In fact, the lamellar microstructures are prone to shearing and globularization under the monotonic deformation [24], [25], [26]. It is of high interest to assess this probability in the case of cyclically deformed microstructures. The latter seems to be theoretically impossible since the shearing and initiation of the globularization process generally necessitate a relatively large amount of strain [27]. However, the present authors come to this idea that the applied reversed path ($R = -1$) in the course of strain-controlled fatigue may influence the mean free path of dislocation movement, dislocation tangling and rearrangement. Accordingly, the shearing of thin β layers may be intensified and the kinetic of early dynamic globularization may in turn be increased. In this regard, Shechtman and Eylon [28] reported the occurrence of the unstable shearing inside the colonies in fatigued microstructure of β-annealed Ti-11 and IMI-685 alloys. Apparently all of these would affect the damage accumulation and in turn the fatigue crack initiation and/or propagation stages.

Coming to the point many aspects of the microstructural evolution of the titanium alloys under the isothermal low cycle fatigue need to be clarified. Accordingly, the present work was conducted to mainly emphasize on the (i) internal damage mechanisms, (ii) substructure development, and (iii) shearing to early globalization during low cycle fatigue of a well-known near-α titanium alloy under the low and high temperature regimes. To this end the dislocation structures and the characters of the interfaces will be precisely examined and compared in heat treated and fatigued microstructures. In this regard, the fatigue fracture characteristics and fatigue properties will be also addressed.

2. Material and experimental procedures

The experimental material was received in hot-forged condition, the chemical composition of which is given in Table 1.

Table 1. The chemical composition of Ti-6242S experimental alloy.

Element (wt.%)	Al	Sn	Zr	Mo	Si	O	Fe	C	N	Ti Balanced
5.9	2.1	4.1	2	0.086	0.03	0.02	0.008	0.01		

The β-transus temperature (Tβ) of Ti6242S near alpha titanium alloy was calculated to be 1005 ± 2 °C [29]. The as-received material was purposefully solution annealed at 50 °C above β-transus temperature for 30 min and then furnace cooled under the controlled rate of approximately 4–6 °C/min. This adapted low cooling rate was considered to provide a proper condition to form "interface phase" layers at α/β interfaces [16]. The heat treated microstructure after the applied heat treatment is shown in Fig. 1. The further details of this fully lamellar structure will be given in the following section.

Fig. 1. The optical image of extra-coarse colony structure of the Ti-6242S experimented alloy after the applied heat treatment.

The uniaxial tensile tests were carried out using a GOTECH AL7000 universal testing machine at room temperature, 200, 400 and 600 °C under the strain rate of 0.001 s^{-1}. The reported tensile properties (yield stress) are the average of three different tests at each condition. The tensile specimens were machined directly from the heat treated material holding the reduced section width of 6 mm and gage length of 30 mm according to ASTM E21 standard [30]. The specimens were first heated up to the test temperature and soaked isothermally for 7 min prior to straining and then subjected to the predetermined tensile tests.

The geometry of isothermal rotary-bending fatigue specimens, which has been machined from the heat treated rods, is shown in Fig. 2.

Fig. 2. Geometry of the thermo-mechanical rotary-bending fatigued specimen. All magnitudes are given in millimeter.

In order to achieve minimum surface defect and avoiding any preferential pre-crack, the specimens' surface were polished down by the finest abrasive grinding paper. The isothermal low cycle rotary-bending fatigue tests were performed at ambient (25 °C) and elevated temperature (200, 400 and 600 °C) by the stress ratio of −1 and rotational speed of 3180 rpm. All of the tests were performed using a fatigue 'ADAMEL LHOMARCY' machine in air and through load-control mode. Considering the tensile properties of the fully lamellar microstructure, the maximum stress was selected to be 500 ± 10 MPa. Tests were carried out at σ_{max}/σ_y ratio, ranging from 0.63 to 1.22 (σ_y is the yield stress of material at each temperature condition as is reported in Table 2). Three specimens were tested for each of the above-mentioned test conditions.

Table 2. Tensile properties and maximum applied cyclic stress to yield stress ratio for fatigue test at different temperature.

Temperature (°C)	Yield stress σ_y (MPa)	Ultimate tensile strength UTS (MPa)	Elongation (%)	(σ_{max}/σ_y)[a]
25 (RT)	783	898	10.2	0.63
200	691	830	11.5	0.72
400	573	722	12.6	0.87
600	408	586	11.7	1.22

a

Maximum applied cyclic stress value is 500 MPa.

Single half of the fatigued specimen was water quenched immediately after fatigue fracture to preserve the deformed microstructure. The fracture surfaces and the microstructures just underneath the fracture tip were studied using scanning electron microscopy (SEM) and transmission electron microscopy (TEM) techniques. Thin foils for TEM analysis were prepared in the closest region possible to the fracture surface (as is depicted in Fig. 2, C-C line). A twin-jet polishing method was employed at the voltage of 35–45 V and the current of 30 mA in the mixed electrolyte of 20 ml perchloric acid, 200 ml methanol and 400 ml n-butanol. For each fatigued specimen, three thin foils were examined therefore a total of ten colonies have been analyzed. The size of prior β grains, width and length of secondary α lamellae and beta layers were measured by a proper image analysis software.

3. Results and discussion

3.1. Before fatigue

The initial microstructure (as heat-treated) holding the arrangements of continuous retained beta layers (β_r) (with average thickness of 500 nm), between the secondary alpha lamellas (α_s) (with average thickness of 2.5 µm) is presented in Fig. 1. This is a typical lamellar colony structure by the average colony size of 190 µm within the prior coarse β grains (average prior β grain size is about 600 µm). For a more detailed view, the TEM images of the initial microstructure are given in Fig. 3.

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Fig. 3. (a) Low-magnification, and (b) high-magnification bright-field TEM images of fully lamellar Ti-6242S experimented alloy before fatigue. The corresponding dark-field TEM image is given in (c); (d) represents the dislocation state at the tip of α lamella. The TEM micrographs have been imaged by FCC [−1 1 0] and BCC [1 1 4] zone axis.

The alternating arrangement of α lamellae and retained β phase is clearly recognized. As is expected, the volume fraction of α-phase is considerably higher than that of β-phase (Fig. 3a). The structure is almost free of dislocations and the dislocation density in both phases is considerably low. The state of dislocation at the tip of α lamellae is precisely analyzed. The presence of stacking faults and straight parallel dislocation lines (planar array of dislocations) clearly indicate planar slip character of α-phase before isothermal fatigue. This can be reasoned considering the low stacking fault energy of α phase in dual phase titanium alloys. The extremely planar slip character of both prism and basal planes was previously reported for near-α titanium alloys holding high aluminum content [31]. It was found that the occurrence of short range ordering (SRO) due to the presence of Ti_3Al precipitates might also increase the planarity of dislocation slip [32]. In Fig. 3b, many alternating thin gray and light lamellae with distinct straight boundaries are also observed within the β phase. All of the internal lamellae are parallel to each other and holding an angle of 40°–60° to the α/β interface. As is well established, the specific contrast of these internal boundaries may imply the presence of local stresses in retained beta phase [33]. The formation of fine lamellae within the β layers was previously ascribed to the precipitation of a body-centered-tetragonal τ phase [34].

Interestingly, the bright and dark field micrographs (Fig. 3b, c) also indicate the presence of a continuous layer with significant contrast by the thickness of about 150 nm at α/β interface. This is known as "interface phase" which has been previously identified as an intermediate transition phase by face-centered cubic (FCC) or HCP structure formed during the β to α transformation [17], [35]. As is observed, the interface phase layers possess sharp and diffused interfaces with β and α phase, respectively. In accordance with the previous findings, these interfacial phases are also symmetric at the two sides of β phase [36]. The corresponding selected area diffraction pattern (Fig. 3b) indicates the FCC crystal structure of the interface phase. Moreover, the interface phase is coherent with the β phase holding an orientation relationship of (−1 −1 1)fcc//(−1 1 0)bcc, 〈0 −1 1〉 fcc// 〈0 0 1〉 bcc along the sharp interfaces. No orientation relationship is distinguished between the interface phase and α phase along the diffused interfaces. The formation of interface phase is mainly attributed to the lattice mismatch between α and β phases imposed through shear mechanism. The interfacial segregation of vanadium at the interface front of β phase and sluggish diffusion of such β stabilizers to the β phase is believed to further promote the formation of this interface layer [14]. In the case of Ti-6242S alloy however, molybdenum by its lower diffusivity in β phase than that of vanadium [37] play an important effect on the interface phase formation. Coming to the point, it is hard to straightforwardly distinguish how such a microstructure may affect the isothermal fatigue strength of the experimented alloy. In this regard, the following sections try to address the influence of various microstructure characteristics on the fatigue damage behavior at different temperatures.

Supplementary, the room temperature (RT) and warm tensile properties of Ti-6242S titanium alloy holding fully lamellar microstructure are also summarized in Table 2. The experimented lamellar coarse-colony microstructure exhibits relatively high strength and ductility at room temperature. As is expected, the yield stress and ultimate tensile strength are decreased by increasing the tensile test temperature, but the elongation to fracture values follow an opposite trend. The maximum applied cyclic stress to yield stress ratio ($\sigma max/\sigma y$) for isothermal fatigue tests have been calculated and the ratios are given in Table 2. Apparently, the higher the ratio, the more severe condition under cyclic loading; this may result in higher tolerated strains through the fatigued specimen.

3.2. After fatigue

3.2.1. Room temperature condition

3.2.1.1. Internal damage mechanism

In order to investigate the microstructure evolution under cyclic loading, the micrographs underneath the fracture surface are examined. In this regard, the cross-sectional view of the fatigued specimen is given in Fig. 4a. The α/β colonies holding different orientations with respect to the loading axis are clearly distinguished. The further details can be well traced in the corresponding TEM images (Fig. 4b–d). The presence of curvature and straight slip lines in α phase (i) and the shearing of β phase and the interface phase in specific shearing directions (the enclosed zone between two dashed lines) (ii) are two of the most important and distinct features of cyclically deformed microstructures. As is observed in Fig. 4b, the β phase and interface phase layers are damaged along the shearing direction, but are not considerably fragmented. However, obvious displacements are recognized between segments of the sheared layers. Surprisingly, a few segments of semi-spherical β phase can be traced through the microstructure (yellow1 circles); this can be considered as a footprint of partial globularization. Considering the scales in which the damage is occurred (about 10 µm), it can be concluded that the shearing is crystallographic and is experienced in specific slip systems [38]. During cyclic loading, the dislocations may pile up within the α phase with high frequency and apply a considerable stress on α/interface phase boundaries barriers. Any break-down of the first barrier enables doubling of the pile-up length thereby the shear band can propagate. The displacements may also cause the dislocations to pile-up in front of the sheared layers.

Fig. 4. (a) Low-magnification cross-sectional SEM micrograph of RT fatigued specimen; (b) and (c, d) are the corresponding low-magnification and high-magnification TEM micrographs, respectively. The TEM micrograph was imaged by FCC [1 1 0] zone axis.

To the best author knowledge, this intensified shearing of the β layers during cyclic loading under the relatively low stress ratio ($\sigma max/\sigma y$), has not been previously reported. However, there are numerous report regarding the shearing of the β phase in near-alpha titanium alloys under the monotonic loading [39], [40], [41]. In the present case, the lamellar microstructure (i), the large colony size (ii) and finally the wavy character of slip (iii) would intensify the occurrence of shearing under the cyclic loading. The latter is attributed to the applied reversed strain path which in turn may increase the dislocation tangling and stress concentration and therefore accelerating the damage accumulation and fatigue cracking. This will be described precisely in the following sections.

3.2.1.2. Substructure development

The substructure development and sub-grain formation in β phase can be clearly traced in the room temperature fatigued microstructures (Fig. 4d). This can be justified considering the easy occurrence of the climb and cross-slip in β phase as a high stacking fault energy phase. As is well established, the sub-grain formation as a thermally activated process tends to take place at relatively lower strain rates, high temperatures and high strain levels [42], [43]. However, in the present case the substructure is well developed in the β phase even at low strain level and ambient temperature. The latter is reasoned considering the short and reversed strain path during cyclic loading, which may well result in dislocation channelization phenomenon within the thin β layers. During cyclic loading, by increasing the number of cycles, the prevalence of tangled structures increases and this in turn may increase the capability of substructure development. In fact, owning to the reversed strain path the dislocations with opposite signs may interact and the annihilation process is intensified.

The low stacking fault energy of α phase and the probable occurrence of short range ordering may restrict the cross-movement of dislocations. As a result, the dislocation structure in this phase generally indicates the predominance of planar slip character, as was observed in the present case before cyclic loading (Fig. 3d). However, by comparison between the two states before and after fatigue, it can be deduced that the slip character changes from planar to wavy. The curvature and straight slip lines in α phase (Fig. 4b) are indicators of mixture wavy/planar slip characters in the room temperature fatigued microstructures. The latter can be justified considering the local increase of the stacking fault energy in α phase. It is believed that the reversed and localized strain path increases the ability of dislocation tangling [44], [45]. This may resemble the condition in which the back stress effect increases thereby putting forces on the partial dislocations and this in turn would decrease the stacking faults widths [46], [47], [48]. This may substantially enhance the crossing ability of moving dislocation. In addition, as the frequency of vacancies is substantial in fatigued structures, the cross-slip capability can be well promoted [49]. The vacancy generation is assumed to occur through edge dislocation dipole trapping and by jog dragging on screw dislocations during cyclic loading [50]. Accordingly, the wavy dislocation configuration within the α phase appears to be reasonable.

In consequence, the substructure development in β phase and the wavy dislocation configuration in α phase may well represent the ability of dislocation tangling, annihilation, rearrangement and subsequently sub-boundary (within the α phase) and sub-grain (within the β phase) formation during room temperature cyclic loading of the experimented alloy.

3.2.1.3. Interface phase effect

The interface phase layers possess a great contribution in the strain accommodation during cyclic deformation. Where the interface phase thickness is negligible compared to the β phase, a greater amount of the strain is accommodated through the β phase; therefore the substructure development and cell formation may be intensified within the β layers. The latter can be clearly traced in fatigue microstructures where small ratios of interface phase to β thickness are found, as is observed in Fig. 4d (dinterface phase/dβ = 0.35). In contrast, if a high ratio of thickness is provided, a considerable part of the imposed strain may be accommodated through the interface phases. Interestingly, the occurrence of deformation twinning within the interface phase layers are clearly observed where the thickness ratio is relatively high, dinterface phase/dβ = 1.27 (Fig. 4c). The corresponding selected area diffraction pattern verifies the occurrence of {1 −1 1}fcc type deformation twins in symmetrical interface phase layers at two sides of the β phase. The latter is introduced as "twinning induced micro-plasticity effect" and is believed to have a positive effect on crack blunting during cyclic loading thereby subsequently improving the fatigue properties. In general, the slip transmission between α and β phases in a fully lamellar structure may difficultly happen, therefore a significant stress concentration is generated behind the α/β interface [17], [45]. This may increase the preferential sites for micro-void formation, coalescence and fatigue crack initiation. The interface phase provides a proper condition for homogeneous strain partitioning through the cyclic loading, and partially compensating the difference in α and β phases plasticity. In fact, the locally concentrated stress caused by dislocation accumulation at the interfaces can be fairly released through the deformation twinning in interface phase layers. The twinning occurrence in interface phase layers with FCC crystal structure and its positive influence on the elongation to fracture values have been previously discussed by Zhao et al. [17]. It is also worth mentioning that the relatively low average thickness of the interface phase layers (~150 nm) may efficiently decrease the effective slip length thereby making the fatigue cracks propagation difficult [51]. This would improve the damaging behavior of the experimented material.

3.2.2. High-temperature condition

3.2.2.1. Internal damage mechanism

The typical microstructures of the specimens which have been fatigue tested at 600 °C, as a representative candidate of high temperature regime, is shown in Fig. 5.

Fig. 5. The TEM micrographs of the specimen which has been fatigued at 600 °C: (a), and (c) represent the shearing of β layers without any considerable displacement; (b) indicates the kink banding in α phase which may cause the shearing and displacement of β layers; and (d) shows the occurrence of stepping in β phase and the presence of geometrically necessary dislocations (GNDs) at α/interface phase boundaries.

As is seen, the shearing and displacement of the β and interface phase layers occurs more extensively owning to the higher strain amplitude. In this context, a large number of segmented layers are observed in the developed microstructures (Fig. 5a, b). In fact, the shearing and displacement of β phase is identified as the main internal damage mechanism at high temperatures and can be interpreted in connection with the occurrence of deformation banding in α phase. The occurrence of banding between β lamellae is indicated by arrow in Fig. 5b. This may cause stepping of the β layers (Fig. 5d) and will then be followed by shearing and fragmentation. The β layers may be sheared without any considerable displacement, and the interface phase left without any significant damage (red dashed circle in Fig. 5a). The β layers can be also completely sheared (yellow dashed circle) or the shearing can be accompanied by displacement (blue dashed circle). These demonstrate that the shearing can be considered as a sequential process. The coherent interface of β/interface phase can't easily pass the dislocations pile-up in comparison to the α/interface phase incoherent boundary. This is attributed to the classic orientation relationship (OR) of the β layers and interface phase, and the large number of coincidence site lattice [52], [53]. It is worth noting that the dislocation pile up is intensified where the deformation band cross at the α/interface phase boundaries. Accordingly, the shearing of the β layers under the present condition can be justified. Comparatively, the room temperature shearing has generally occurred perpendicular to the β phase layers. However, at higher temperature condition the shearing may also be found at various angles with respect to the β layers (Fig. 5c). This may be attributed to the activation of secondary slip systems at higher temperatures in the α phase.

The micrographs of cross sectional view underneath the fracture surface were further analyzed to clarify the influence of the internal damage mechanism on the crack initiation and propagation. As is observed in Fig. 6a, shearing of the β layers is mainly restricted by the colony size. In addition, the main fatigue crack initiates where the localized shear is intensified due to the specific orientation of the colonies relative to the stress axis.

Fig. 6. Cross sectional micrographs of the fatigued specimen at 600 °C, imaged by SEM: (a) fatigue crack initiation, and (b) propagation due to intensification of localized shear.

The intensified localized shear regions can be clearly traced along and at the vicinity of the main crack propagation path (Fig. 6b). The shearing and displacement of the β layers may also result in initiation and propagation of the micro-cracks, as are identified below the fracture surface of the fatigued specimens (Fig. 7). Finally, it can be concluded that the damage mechanisms in fully lamellar microstructure comprising interface phase layers under cyclic loading at ambient and elevated temperatures is restricted to the shearing of the β layers. As is observed, there is no cavity along α/interface phase boundaries in fatigued microstructures. This is attributed to the positive influence of the interface phase layers on the strain accommodation between secondary alpha phase and retained beta layers. Where the cyclic deformation doesn't cause stepping or fragmentation, the geometrically necessary dislocations (GNDs) are formed at α/interface phase boundaries to maintain the strain compatibility (Fig. 5d).

Fig. 7. SEM micrographs of cross-sectional view of fatigued specimen at 600 °C (white line and gray area represent the beta phase and alpha phase, respectively): (a) microstructural evolution underneath the fracture surface; (b) and (c) magnified view of marked area in (a).

3.2.2.2. Initial stage of dynamic globularization

As is logically deducted, the intensity of strain localization can be increased due to the applied reversed strain path (R = −1). Owning to this comprehensive unstable shearing the microstructure alters to a wavy-like structure containing kinked or sheared β layers (Fig. 7). The shearing and displacement mechanisms would lead to a considerable segmentation of the β layers. These segments with various lengths can be found through the fatigued microstructures (Fig. 7c). This is more intensified at higher temperatures due to the relatively higher strain amplitude. The fragmentation is also found to be significantly influenced by the specific orientation of α/β colonies with respect to loading axis. This in turn may significantly reduce the aspect ratio of the β layers (Fig. 7c). As is seen in Fig. 8a, the segmentation of the β layers may be associated with localized shear deformation and/or boundary splitting.

Fig. 8. TEM micrographs of isothermally fatigued specimen at 600 °C: (a) two type of segmentation of the β phase during cyclic deformation; β fragmentation along the interior sub-boundary and shearing of the β layer, and (b) the initial stage of early dynamic globularization, the penetration of the β phase along the α phase sub-boundary, the arrows indicate the β grooves.

The boundary splitting mechanism requires the presence of internal boundaries within the lamellar phases. The boundary may be a pre-existing grain boundary, a sub-grain boundary, or any deformation band induced through the imposed cyclic strain [54]. The micro-mechanism of the β phase fragmentation via boundary splitting/grooving is associated with (i) the generation of transverse boundaries within β layer, (ii) the subsequent formation of grooves associated with the instability of 90° dihedral angles between β/interface phase and β/β boundaries (red arrow in Fig. 8b indicates the grooves within the β layers), and (iii) complete lamellar fragmentation by deepening of the grooves from both sides of the β layer (Fig. 8b). It should be noted that the thinner β layers are more readily divided into the separated fragments by boundary splitting along the transverse boundaries [55], [56].

Interestingly, the early dynamic globularization of α lamellae may also be found through the high temperature fatigued microstructures. The involved micro-mechanism appears to be sub-boundary induced splitting and/or localized shear induced splitting across α lamellae along with the subsequent diffusional penetration of the β phase along the boundaries. This is clearly seen in Fig. 8b. The main factor inhibiting the globularization via boundary splitting is most likely related to the greater thickness of α lamellae [54]. Generally, the lamella thickness, deformation temperature, and the nature of the interfaces (e.g. boundary energy) are the main factors influencing the extent and kinetics of early globularization. In this regard the occurrence of dynamic recovery and shear band formation within α lamellae may strongly affect the fragmentation and subsequently the early globularization. In addition, the globularization of the lamellar structure can be enhanced considerably by the conversion of the initially semi-coherent boundaries/interfaces into the higher-energy non-coherent ones; and/or the creation of interfaces with relatively low coherency during heat treatment [57], [58]. Therefore, the α/interface phase boundaries as a high-energy non-coherent interface without any classic orientation relationship may promote the boundaries splitting/grooving process and consequently influence on the strain level required for dynamic globularization.

It is worth mentioning that the lamellar structure is generally very stable for a fixed strain path during warm deformation [24], [59]. In addition, by decreasing the deformation temperature the strain level required for dynamic globularization is significantly increased [23], [54]. However, in the present case the change of strain path may enhance the rate of globularization even under the room temperature condition. In fact, the microstructures under fatigue, due to applying the cyclic deformation with reversal strain path are capable of substructure development. This provides a proper condition for sub-boundary diffusion, and therefore the microstructure may experience the dynamic globularization earlier than that of monotonic deformation.

3.2.3. Fatigue properties

The isothermal low cycle fatigue behavior of the experimented Ti-6242S alloy is displayed in Fig. 9. All of the fatigue tests were carried out at a stress amplitude of 500 MPa. So, increase in temperature is accompanied by increase in maximum applied cyclic stress to yield stress ratio ($\sigma max/\sigma y$). The cyclic deformation behavior is found to be dependent strongly upon the fatigue temperature. As is realized, the number of cycles to failure (Nf) is decreased by raising the fatigue temperature.

Fig. 9. Temperature vs. number of cycles to failure obtained from the rotary-bending isothermal fatigue tests.

The macroscopic and microscopic fractographs of the fatigue fracture surface at room temperature and 600 °C were analyzed in detail, and are given in Fig. 10, Fig. 11. The fatigue fracture surfaces can be divided into three regions: (i) fatigue crack initiation, (ii) fatigue crack propagation, and (iii) final fracture regions [10], [60], [61], [62].

Fig. 10. The macroscopic fracture surface and high magnification SEM fractographs of Ti-6242S alloy which were fatigue tested at (a) room temperature condition, and (b) high temperature condition (600 °C). Small rectangles in (b) display multiple initiation sites.

Fig. 11. The fatigue striations in stable crack propagation region on the fracture surface of the specimen fatigued at: (a) low temperature condition (RT), and (b) high temperature condition (600 °C).

The large and relatively smooth areas can be clearly traced in crack propagation region for the both temperature regimes. There are also a large number of river patterns; following the reverse spreading direction of them gives the fatigue cracks initiation sites (indicated by the red rectangles in Fig. 10a and b). In contrast, the final fracture regions represent a ductile mode of fracture, where a mixture of deep and shallow dimples is recognized by higher magnification. As is expected, the average dimples diameter is larger at higher fatigue temperature. At room temperature the fatigue failure occurs through one predominant surface crack initiation, but at higher temperatures the multiple initiation sites are observed and the final fracture stage occurs approximately at the center of fatigue fracture surface (Fig. 10b). The fatigue striations in stable crack propagation region are also clearly visible (Fig. 11a and b); each striation is linked by a single stress cycle and represents the successive progress of the fatigue crack propagation [63], [64].

As is seen in Fig. 11, increasing the test temperature has resulted in greater fatigue striations' spacing; this may indicate the higher fatigue crack growth rate. The latter is basically attributed to the increase in number of active slip systems and their contributions during cyclic loading [65]. In the present case, the shearing and displacement of the lamellar phases at high temperature condition can also be influential. The multiple crack initiation sites and the facilitated propagation of the main crack may well describe the lower fatigue life at higher temperatures and maximum applied cyclic stress to yield stress ratio (σ_{max}/σ_y). Obviously, the fatigue behavior of the experimented material is completely coincided with the previous described microstructural evolution.

4. Conclusion

Evolution of the fully lamellar microstructure in Ti-6242S alloy during isothermal low cycle fatigue at room and high temperatures was investigated. The microstructures were significantly affected by the characteristic of various interface boundaries. The main conclusions are summarized as follows:

(1)

The shearing and displacement of the β layers were identified as dominant mechanisms of internal damage during isothermal fatigue. This was intensified at higher fatigue temperature owning to the intensified slip localization.

(2)

The substructure development and cell formation were characterized within the β phase during room temperature cyclic loading. This was attributed to the dislocation channelization phenomenon as a result of the reversed and localized strain path. Interestingly, the α phase represented mixed wavy/planar slip character which promoted the sub-boundary formation. This was also justified considering local variation of the staking fault energy owning to the back stress effect during cyclic loading.

(3)

The α/β interface phase rendered a considerable effect on the damage behavior through twinning induced micro-plasticity effect and homogenizing the strain partitioning.

(4)

The initial stage of dynamic globularization during cyclic loading at low temperature and imposed strain was characterized. The globularization kinetic of the β layers was intensified at higher temperature and the sub-boundary induced boundary splitting/grooving was introduced as an involved mechanism. Interestingly, the early dynamic globularization of the α lamellae had also occurred via a micro-mechanism based on the sub-boundary and/or localized shear induced splitting across α lamellae and the diffusional penetration of the β phase along the boundaries through the high temperature fatigued microstructures.

(5)

Fractography of the fatigued specimens demonstrated that the fatigue crack initiation and propagation were significantly affected by fatigue temperature. A larger striations width was correlated with multiplication of the slip systems and acceleration of fatigue crack propagation due to shearing and displacement by raising the fatigue temperature.

References