Cyclic deformation and microstructure evolution of high Nb containing TiAl alloy during high temperature low cycle fatigue

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Abstract

The cyclic stress-strain (CSS) behavior of high Nb containing TiAl alloy with a duplex microstructure was investigated at 850 ° C. Transmission electron microscope (TEM) observations and electron backscattered diffraction (EBSD) techniques were carried out to obtain insight into the microstructure evolution governing this behavior. At low strain amplitude, the material exhibits a rapid saturation of stress amplitude. At intermediate and high strain amplitude, the CSS behavior is characterized by generally cyclic softening. The changes of microstructure are strain-induced phase transformations and dynamic recrystallization, which lead to a degradation of lamellar microstructure. Twin boundaries can promote discontinuous dynamic recrystallization of γ phase, for processing relatively high energy. $\alpha_2 + \gamma \rightarrow B2$, $\alpha_2 \rightarrow B2$, and $\alpha_{2\text{lamellae}} \rightarrow \gamma$ phase transformations are detrimental to fatigue life of the material, because fracture propagation are usually along B2 phase boundaries and γ grain boundaries, thus, the fracture mode is a combination of ductile fracture and intergranular cleavage fracture..

Keywords: Cyclic stress-strain behavior; Low cycle fatigue; High Nb containing TiAl alloy; Recrystallization

1. Introduction

TiAl-based alloys have received considerable attention in recent years, already being used in some areas, due to their excellent high temperature properties and low density [1-5]. At the end of last century, the third generation γ -TiAl alloys containing high Nb addition were designed, which improves mechanical properties, oxidation and creep resistance, compared with the previous generation of TiAl-based alloys [6-9].

In the most foreseen applications, as an engineering material used at high temperature, TiAl alloys are subjected to fluctuating or cyclic loading during service, which give rise to plenty of plastic deformation [6, 10, 11]. Thus, low cycle fatigue tests are used to evaluate the lifetime of TiAl-based alloys, simulating service conditions. There are many studies on TiAl alloys regarding

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low cycle fatigue. It is known that poly-synthetically twinned (PST) crystals are a model material with a perfectly defined lamellae orientation, respecting to loading axis, during low cycle fatigue tests, and Umakoshi et al. [10, 12] have clearly identified the deformation mechanisms of PST crystals, governing the CSS behavior at room temperature and 750° C. A. L. Gloanec and co-workers [13, 14] have investigated the low cycle fatigue cyclic deformation mechanisms and substructures of Ti-48Al-2Cr-2Nb alloy with a nearly fully lamellar microstructure. However, as the third generation of TiAl-based alloys, high Nb-containing alloys are scarcely studied, giving insight into low cycle fatigue investigations. Fritz Appel et al. [11] observed microscopic characterization of Ti-45Al-8Nb-0.2C alloy with a lamellar microstructure at room temperature, 550° C and 850° C. T. Kruml et al. [15] investigated the low cycle fatigue properties of a nearly lamellar Ti-48Al-7.8Nb-0.2Ni alloy, compared with TiAl alloys with 2at% Nb. Nevertheless, the question remains as to whether these properties and microstructures also hold for a duplex microstructure of high Nb-containing alloy, which consists of the fine lamellar structure of the γ and α_2 phases and the equiaxed γ phase. Additionally, because the high-Nb containing alloys are used at temperatures 60-100° C higher than those with 2at% Nb, the low cycle fatigue properties at the service temperatures need to be investigated. To this end, low cycle fatigue tests are performed at 850° C on a Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy with a duplex microstructure, coupled with TEM observations and EBSD techniques in the present study.

2. Experimental procedure

High Nb containing TiAl forged material was investigated in the present study, with a nominal composition of Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y (at. %), and the microstructure is shown in Fig. 1. The specimen geometry is shown in Fig. 2. As shown, the cylindrical fatigue samples with a gauge section of 6 mm in diameter and 32 mm in length and a total length of 90 mm were machined. The final surface of the tested pieces was achieved by mechanical polishing, using SiC papers, to minimize the effects of irregularities on fatigue lives. Cyclic deformation tests were conducted on an MTS 810 servohydraulic machine at 850° C with different total strain amplitudes and a constant strain rate of $5 \times 10^{-3} \text{ s}^{-1}$ in laboratory air. The signal is triangular, and the loading cycle is symmetrical ($R_\epsilon = \epsilon_{\min} / \epsilon_{\max} = -1$).

The microstructures of low cycle fatigue tests were examined using JEM 2010 transmission electron microscope (TEM) operated at 300kV, backscattered electron microscopy (SEM-BSE), and electron backscattered diffraction (EBSD). The thin foils for TEM were prepared from the gauge length of the samples near the fracture surface and cut perpendicularly to the loading axis. By mechanical grinding, the thickness of the foils was reduced to approximately 70 μm and then prepared by a twin-jet polisher. For EBSD observations, 3 mm (in height) samples were cut parallel and next to the TEM foils, the sample surfaces oriented at an angle of 70° to the horizontal, with a step size of 0.2 μm . The accumulated EBSD data were post-treated with Channel 5TM software.

3. Results and discussion

3.1 Cyclic stress-strain behavior

Fig. 3a presents the cyclic stress-strain (CSS) behavior of the tested TiAl alloy, i.e., the variation of the stress amplitude $\Delta\sigma/2$ measured as a function of the number of cycles at 850 ° C. Accordingly, the material exhibited a rapid saturation at low $\Delta\varepsilon_f/2$ values ($\Delta\varepsilon_f/2=\pm 0.32\%$) and a significant cyclic softening at high $\Delta\varepsilon_f/2$ values ($\Delta\varepsilon_f/2=\pm 0.70\%$). Besides, at $\Delta\varepsilon_f/2=\pm 0.40\%$, the material presented cyclic softening after a slight initial hardening about 150 cycles. At $\Delta\varepsilon_f/2=\pm 0.32\%$, the specimen did not fail until 10^4 cycles, while the fatigue life rapidly decreased at $\Delta\varepsilon_f/2=\pm 0.70\%$. Fig. 3b shows the plastic deformation accumulated in a cycle of the high-Nb containing TiAl alloy. The plastic strain amplitude $\Delta\varepsilon_p$ was stable at $\Delta\varepsilon_f/2=\pm 0.32\%$ and increased at $\Delta\varepsilon_f/2=\pm 0.40\%$ and $\Delta\varepsilon_f/2=\pm 0.70\%$. It is indicated that, at $\Delta\varepsilon_f/2=\pm 0.32\%$, the material exhibits cyclic stability, but cyclic softening at $\Delta\varepsilon_f/2=\pm 0.40\%$ and $\Delta\varepsilon_f/2=\pm 0.70\%$. The increase of plastic strain amplitudes seems to be detrimental, because the fatigue life is reduced when the strain amplitude increases. TiAl alloys with 2 at. % Nb or 2Nb-2Cr were investigated at the cyclic temperature of 850 ° C, for the intrinsic limitation of the alloys. Moreover, plenty of studies have shown that cyclic hardening typically occurred in these alloys at room temperature [14, 16-18]. It has been shown that the cyclic deformation can be produced by dislocation glide and twinning, which depend on the applied strain amplitude. F. Appel and co-workers [11] were investigated on TVB-2 (Ti-45Al-8Nb-0.2C) alloy and observed cyclic softening at 850 ° C with total strain amplitude of $\Delta\varepsilon_f/2=\pm 0.70\%$, which was similar with the result in the present study.

Table 1 Summary of low cycle fatigue data of the Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy at R=-1 and 850 ° C.

$\Delta\varepsilon_f/2$ (%)	$\Delta\varepsilon_p/2$ (%)	N_f
0.32	1.77×10^{-2}	11592
0.40	6.25×10^{-2}	1927
0.70	0.28	115

$\Delta\varepsilon_f/2$ – total strain amplitude, $\Delta\varepsilon_p/2$ – accumulated plastic strain, N_f – number of cycles to failure.

3.2 Phase transformation in cyclic deformation

The changes in microstructure after low cycle fatigue tests were observed at various strain amplitudes. It can be seen that the changes of internal structures are detected at low $\Delta\varepsilon_f/2$ value ($\Delta\varepsilon_f/2=\pm 0.32\%$), intermediate $\Delta\varepsilon_f/2$ value ($\Delta\varepsilon_f/2=\pm 0.40\%$) and high $\Delta\varepsilon_f/2$ value ($\Delta\varepsilon_f/2=\pm 0.70\%$), as shown in Fig. 4. The lamellar microstructure is partially replaced by γ phase in all strain levels investigated in this study, resulting from the gradual disappearance of α_2 lamellae (Fig. 5), and can be associated with $\alpha_{2\text{lamellae}} \rightarrow \gamma$ phase transformation during low cycle fatigue deformation at 850 ° C. Inside the γ phase, at total strain amplitude of $\Delta\varepsilon_f/2=\pm 0.32\%$, the dislocation density is low, dislocations piled-up at the interface of γ phase and lamellae, while at an intermediate strain amplitude of $\Delta\varepsilon_f/2=\pm 0.40\%$, the dislocation density is relatively high. Alternatively, stacking faults can be found, which might be resulted from the compatible deformation of $\beta/B2$ phase between γ and α_2 phase during the cyclic deformation. γ , α_2 and B2 (ω) phases were confirmed by their selected area diffraction patterns (SADPs) shown in Fig. 6b, c and d, respectively. At high strain amplitude of $\Delta\varepsilon_f/2=\pm 0.70\%$, dislocations are widely spread and tangled in the γ phase.

Fig. 6 displays the TEM microstructures after fatigue with total strain amplitude of $\Delta\varepsilon_f/2=\pm 0.40\%$. The B2 precipitates touched both α_2 and γ blocks, indicating that it grew by consuming both α_2 and γ phases, that is, this process is mainly associated with $\alpha_2 + \gamma \rightarrow B2$

phase transformation. Interestingly, the B2 block has a partly precipitated ω phase. The existence of ω phase was revealed by the SADPs shown in Fig. 6c and d. According to the quasi-binary phase diagram with high Nb addition [19], the ω phase can be a equilibrium phase at temperatures of 800-1000 °C. It is necessary to investigate the ω phase not only for the temperature range is expected to be the application temperature range but the areas of B2 phase with ω phase coexisting are commonly observed to be the nucleation sites of cracks [19-21]. Many researchers have focused on β /B2 (ω) phase transformation [22-24], Z.W. Huang et al. [24] found that there was no independent B2 phase in Ti-44Al-8Nb-1B alloy after being exposed to 700 °C air for up to 3000 h, and this finding implies that the formation of the B2 phase is accompanied with the formation of the ω phase. Therefore, the ω phase might be removed during cyclic deformation at 850 °C, but the reason is not clear.

Fig. 7 presents the phase distribution maps of samples 1-6. It should be mentioned that sample 1 is the forged Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy, which was cut 6 mm in diameter and 10 mm in length, heat treated for 8h15min, 4h43min for sample 2 and 10min for sample 3 at 850 °C and then followed air cooling according to the low cycle fatigue tests. Comparing sample 1 to sample 4, sample 2 to sample 5, and sample 3 to sample 6, the volume fractions of γ phase were decreased and β /B2 phase were increased. It indicates that the effect of strain amplitude seems detrimental on fatigue life of the material, because B2 phase fractions are increased after fatigue tests, and the blocks of B2 phase could lead to easier crack propagation and cause earlier failure. The volume fractions of α_2 phase were increased after cyclic deformed with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$ and of $\Delta\epsilon_f/2=\pm 0.70\%$, but decreased with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$. In the present study, regarding to the temperature of low cycle fatigue tests (850 °C), there exists ($\alpha_2 + \gamma + \beta$ /B2) phase field and ω is an equilibrium phase [25]. Therefore, it can occur $\alpha_2 + \gamma \rightarrow \text{B2}$, $\alpha_2 \rightarrow \text{B2}$, and $\gamma \rightarrow \alpha_2$ phase transformations. These processes can be confirmed from Fig. 8. α_2 block in grey contrast was precipitated within the γ phase, β /B2 (ω) particles in white contrast were surrounding the precipitated α_2 block. This was resulted from $\gamma \rightarrow \alpha_2$ phase transformation and then followed $\alpha_2 + \gamma \rightarrow \text{B2}$ and $\alpha_2 \rightarrow \text{B2}$ phase transformations. To increase the accuracy of the phase distribution data for its statistical distribution, six mappings are conducted for each sample, and the data listed here are the mean values.

3.3 Dynamic recrystallization in cyclic deformation

In the specimen subjected to low cycle fatigue tests and failed with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$, the microstructure appeared areas of γ phase islands, in which tangled high-density dislocations and grew across many lamellae during the following deformation (Fig. 9a). Furthermore, intensive twinning were observed, which did not across the lamellar boundaries and were adjacent to the recrystallized γ phase. It seems that the recrystallized γ phase was relieved stress concentration caused by twinning and avoided the occurrence of cracks. At total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$, shown in Fig. 9b, a similar recrystallized γ phase can be found, however, the substructure was different. Not only dislocations can be found but also twinning. Fig. 9c shows that at total strain amplitude of $\Delta\epsilon_f/2=\pm 0.70\%$, lamellar structures were replaced by single γ phase with high-dislocation density. Indeed, this replacement was in a substantial part of volume in the material. It should be note that there were also different directions of twins in the γ grain, and at the intersection of these twins, γ phase was recrystallized.

The observations reported here indicate that the changes of microstructure are strain-induced phase transformations and dynamic recrystallization. The deformation mechanisms, namely dislocation glide and twinning, are the same in PST crystals and a duplex microstructure of high Nb containing TiAl alloy. Many researchers have engaged in studying the mechanisms of TiAl alloys during cyclic deformation [11, 14, 26, 27]. The perfect dislocations present in γ phase have the Burgers vectors $b_{\langle 110 \rangle} = 1/2\langle 110 \rangle$, $b_{\langle 011 \rangle} = \langle 011 \rangle$, and $b_{\langle 112 \rangle} = 1/2\langle 112 \rangle$ and preferentially glide on $\{111\}_{\gamma}$ planes. The superdislocations have the Burgers vectors $b_{\langle 011 \rangle}$ and $b_{\langle 112 \rangle}$, and its dissociation reactions are various, involving different planar faults. Mechanical twins are along $1/6\langle 11\bar{2} \rangle\{111\}$. However, the mechanical performances are not only affected by the dislocation and twinning mechanisms, but the synergetic interactions of the individual constituents. The investigation of these processes, i. e. phase transformations and dynamic recrystallization, by TEM and EBSD observations are the subject matter of this paper.

Fig. 10 shows misorientation angle distributions of the γ phase in samples 1-6. Obviously, there are two types of small peaks around the misorientation angle of $60\pm 3^{\circ}$ and $70\pm 3^{\circ}$, which are mainly related to the pseudo twin boundaries, namely $60^{\circ}\langle 111 \rangle$, and the true twin boundaries, namely $70^{\circ}\langle 110 \rangle$, according to the work by Jin et al. [28]. It should be noted that $70^{\circ}\langle 110 \rangle$ twin boundaries are equivalent to $180^{\circ}\langle 111 \rangle$ twin boundaries in the γ phase. Besides, there is a dominant misorientation angle of $89\pm 3^{\circ}$ in all the six samples. Investigations [22, 29, 30] showed that this peak was associated with dynamic recrystallization (DRX) grains. Two features of the DRX grains were found, the first one is DRX grains were located within the matrix grains and the other one is their matrix grains were DRX grains or underwent large deformation. Frequencies of $60\pm 3^{\circ}$, $70\pm 3^{\circ}$, and $89\pm 3^{\circ}$ misorientation angle were summarized in Table 2. Fractions of misorientation angle $60\pm 3^{\circ}$ and $70\pm 3^{\circ}$ are decreased during the low cycle fatigue tests, comparing sample 1 to sample 4, sample 2 to sample 5, and sample 3 to sample 6. Decreasing of the twin boundaries may be resulted from the above-described phase transformations and dynamic recrystallization, because γ phase has crystal orientations ($L1_0$ crystal structure with lattice parameters $a = b \neq c$) for which dislocation glide and mechanical twinning are difficult, and phase transformation can change the lattice parameters and the elastic constants, which accommodates constraint stress between neighbored grains. Besides, dynamic recrystallization may also decrease the fractions of twin boundaries, which can be confirmed in Fig. 9c.

Table 2 Summary of $60\pm 3^{\circ}$, $70\pm 3^{\circ}$, and $89\pm 3^{\circ}$ Misorientation angle fractions

Sample	$60\pm 3^{\circ}$ (%)	$70\pm 3^{\circ}$ (%)	$89\pm 3^{\circ}$ (%)
1	9.09	6.55	73.15
2	8.18	7.34	65.41
3	10.13	6.75	62.45
4	4.63	3.37	81.89
5	5.86	3.14	79.08
6	6.51	3.99	79.20

The γ phase, according to X.P. Song and G.L. Chen ever reported [31], possesses low stacking fault energy (80 ± 5 mJ/m²), which can readily experiences discontinuous dynamic recrystallization (DDRX). This DDRX process is mainly comprises grain nucleation and growth processes. No matter what type of twin boundaries, they possess a relatively high energy that can

supply nucleation sites for DDRX and result in the different free energy between the parent and the product grains, which can promote the recrystallized grains to grow up. The following steps are followed by Tango of Channel 5TM software during the evaluation of deformed, substructure and recrystallized fractions:

1. First the Tango does a grain reconstruction (using the user defined parameters in the grain area determination window).
2. Then it takes each grain and measures the internal average misorientation angle within the grain.
3. If the average angle in a grain exceeds the user-defined minimum angle to define a subgrain, (θ_c), the grain is classified as being “deformed”.
4. Some grains consist of subgrains whose internal misorientation is under θ_c but the misorientation from subgrain to subgrain is above θ_c . In that case the grain is classed as “substructure”.
5. All the remaining grains are classified as “recrystallized”.

Where θ_c is set as 2° and θ_{GB} is set as 15° in the paper.

As shown in Fig. 11e and Table 3, recrystallized and substructure fractions were reduced whereas deformed fractions were increased comparing sample 1 to sample 4 and sample 2 to sample 5. Recrystallized and deformed fractions were reduced while substructure fractions were increased comparing sample 3 to sample 6. It is indicate that the total strain amplitudes can reduce the recrystallized fractions, regarding as the same to plastic strain amplitudes. But at relatively low strain amplitudes ($\Delta\epsilon_f/2=\pm 0.32\%$) and medium strain amplitudes ($\Delta\epsilon_f/2=\pm 0.40\%$), the substructure fractions were reduced and deformed fractions were increased, which was contrary to the results of high strain amplitudes ($\Delta\epsilon_f/2=\pm 0.70\%$). It seems that at total strain amplitude of $\Delta\epsilon_f/2=\pm 0.70\%$, new dislocation slip systems are hard to activate and mechanical twinning may difficult to occur, so sub-grain boundaries are increased to release constraint stresses. Besides, the fractions of sample 1, 2, and 3 showed that the effect of time that the material underwent at 850°C on recrystallized fraction was not significant, however, the substructure fraction was increased and deformation fraction was reduced with increasing the heat-treatment time. This was related to recovery processes.

Table 3 Volume fractions of recrystallized, substructure and deformed grains in samples 1-6.

Sample	Recrystallized (%)	Substructure (%)	Deformed (%)
1	29.51	49.94	20.55
2	30.25	46.29	23.46
3	29.65	42.39	27.96
4	21.53	43.42	35.05
5	25.85	45.28	28.87
6	26.23	48.05	25.72

3.4 Fracture behavior

Fig. 12 shows the fracture surfaces of the high-Nb containing TiAl samples cyclic deformed at 850°C , with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$, $\Delta\epsilon_f/2=\pm 0.40\%$, and $\Delta\epsilon_f/2=\pm 0.70\%$, respectively. The fracture surface was smooth and some secondary cracks were found

perpendicularly from the fractured surfaces. At total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$, intergranular cleavage and some dimples were observed (Fig. 12a). Secondary cracks were widely found at $\Delta\epsilon_f/2=\pm 0.40\%$ and the dimples became more and larger because of high strain concentration at total strain amplitude of $\Delta\epsilon_f/2=\pm 0.70\%$. It indicates that the fracture mode of the alloy cyclic deformed at 850°C is a combination of ductile fracture and intergranular cleavage fracture.

Fig. 13 shows the crack propagation near fracture surfaces of the low cycle fatigue samples. At total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$, the fracture was propagated along γ grain boundaries and $\beta/\text{B2}$ phase boundaries. Some interlamellar cracks were also observed. At total strain amplitude of $\Delta\epsilon_f/2=\pm 0.70\%$, formation of intensive pores were occurred during deformation, and the pore elongation was also along γ grain boundaries and $\beta/\text{B2}$ phase boundaries. Therefore, it seems that $\alpha_2 + \gamma \rightarrow \text{B2}$, $\alpha_2 \rightarrow \text{B2}$, and $\alpha_{2\text{lamellae}} \rightarrow \gamma$ phase transformations and γ phase recrystallization were detrimental to low cycle fatigue of TiAl alloy at 850°C , which could reduce fatigue life of the material.

4. Conclusions

The cyclic stress-strain (CSS) behavior and microstructure evolution of duplex high Nb containing TiAl alloy were investigated at the temperature of 850°C with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$, $\Delta\epsilon_f/2=\pm 0.40\%$, and $\Delta\epsilon_f/2=\pm 0.70\%$. The following conclusions are drawn:

- 1) The cyclic stress-strain (CSS) behavior of Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y, with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$, exhibits a rapid saturation of stress amplitude. At total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$ and $\Delta\epsilon_f/2=\pm 0.70\%$, the material generally presents cyclic softening.
- 2) Phase transformations and dynamic recrystallization lead to a degradation of lamellar microstructure. $\alpha_2 + \gamma \rightarrow \text{B2}$, $\alpha_2 \rightarrow \text{B2}$, and $\alpha_{2\text{lamellae}} \rightarrow \gamma$ phase transformations are detrimental to fatigue life, because fracture propagation are usually along B2 phase boundaries and γ grain boundaries. The fracture mode of the alloy cyclic deformed at 850°C is a combination of ductile fracture and intergranular cleavage fracture.
- 3) $60^\circ\langle 111 \rangle$ pseudo and $70^\circ\langle 110 \rangle$ true twin boundaries of γ phase can promote DDRX process and the fractions of twin boundaries are decreased during cyclic deformation.

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Figure Legend

- Fig. 1 Duplex microstructure of Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy imaged by scanning electron microscopy.
- Fig. 2 Geometry of the fatigue specimen.
- Fig. 3 Duplex Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy cyclic deformed at 850 °C: a. cyclic stress-strain behavior, b. accumulated plastic strain in a cycle, c. magnified cyclic stress-strain behavior of $\Delta\epsilon_f/2=\pm 0.70\%$, d. magnified plastic strain accumulated in a cycle of $\Delta\epsilon_f/2=\pm 0.70\%$.
- Fig. 4 Partially destroyed lamellar microstructure fatigued at 850 °C: a. low dislocation density in the γ phase after N=11592 cycles with R=-1 and a total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$. b. stacking faults and high dislocation density in the γ phase after N=1927 cycles with R=-1 and a total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$. c. high dislocation density in the γ phase after N=115 cycles with R=-1 and a total strain amplitude of $\Delta\epsilon_f/2=\pm 0.70\%$. Foil planes were perpendicular to the loading axis.
- Fig. 5 The front of the resting lamellar microstructure in γ phase. 850 °C, $\Delta\epsilon_f/2=\pm 0.32\%$, $\Delta\sigma_f/2=386$ MPa, foil plane was perpendicular to the loading axis, N=115 cycles.
- Fig. 6 a. TEM bright field observation of low cycle fatigue microstructure at 850 °C: total strain $\Delta\epsilon_f/2=\pm 0.40\%$, $\Delta\sigma_f/2=\pm 431$ MPa, foil plane was perpendicular to the loading axis, N=1927 cycles. Selected area diffraction patterns (SADPs) of α_2 (b), B2 (ω) (c), ω (d) phases.
- Fig. 7 Phase distribution maps of the samples: a. sample 1, heat-treated for 8h15min at 850 °C. b. sample 2, heat-treated for 4h43min at 850 °C. c. sample 3, heat-treated for 10min at 850 °C. d. sample 4, fatigue with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$, $\Delta\sigma_f/2=\pm 386$ MPa, and failed after N=115 cycles. e. sample 5, fatigue with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$, $\Delta\sigma_f/2=\pm 431$ MPa, and failed after N=1927 cycles. f. sample 6, fatigue with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.70\%$, $\Delta\sigma_f/2=\pm 538$ MPa, and failed after N=11592 cycles. Sample planes were perpendicular to the loading axis, fatigue at 850 °C, R=-1.
- Fig. 8 Microstructure of Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy after fatigue at 850 °C with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$.
- Fig. 9 Strain induced phase transformation after fatigue at 850 °C: a. γ phase recrystallized in lamellae and plenty of dislocations formed in during the following deformation at $\Delta\epsilon_f/2=\pm 0.32\%$, $\Delta\sigma_f/2=\pm 386$ MPa, foil plane was perpendicular to the loading axis, N=115 cycles. b. γ phase recrystallized in lamellae and twinning occurred during the deformation thereafter at $\Delta\epsilon_f/2=\pm 0.40\%$, $\Delta\sigma_f/2=\pm 431$ MPa, foil plane was perpendicular to the loading axis, N=1927 cycles. c. recrystallization of γ phase occurred at the intersection of twins at $\Delta\epsilon_f/2=\pm 0.70\%$, $\Delta\sigma_f/2=\pm 538$ MPa, foil plane was perpendicular to the loading axis, N=11592 cycles.
- Fig. 10 Misorientation angle distribution of γ phase in the samples: a. sample 1, heat-treated for 8h15min at 850 °C. b. sample 2, heat-treated for 4h43min at 850 °C. c. sample 3, heat-treated for 10min at 850 °C. d. sample 4, fatigue with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.32\%$, $\Delta\sigma_f/2=\pm 386$ MPa, and failed after N=115 cycles. e. sample 5, fatigue with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.40\%$, $\Delta\sigma_f/2=\pm 431$ MPa, and failed after N=1927 cycles. f. sample 6, fatigue with total strain amplitude of $\Delta\epsilon_f/2=\pm 0.70\%$, $\Delta\sigma_f/2=\pm 538$ MPa, and failed after N=11592 cycles.

- Fig. 11 Recrystallized maps of the samples: a. sample 1, heat-treated for 8h15min at 850 °C. b. sample 2, heat-treated for 4h43min at 850 °C. c. sample 3, heat-treated for 10min at 850 °C. d. sample 4, fatigue with total strain amplitude of $\Delta\epsilon_t/2=\pm 0.32\%$, $\Delta\sigma_t/2=\pm 386$ MPa, and failed after N=115 cycles. e. sample 3, fatigue with total strain amplitude of $\Delta\epsilon_t/2=\pm 0.40\%$, $\Delta\sigma_t/2=\pm 431$ MPa, and failed after N=1927 cycles. f. sample 4, fatigue with total strain amplitude of $\Delta\epsilon_t/2=\pm 0.70\%$, $\Delta\sigma_t/2=\pm 538$ MPa, and failed after N=11592 cycles. g. volume fractions of recrystallization of samples 1-6. Sample planes were perpendicular to the loading axis, fatigue at 850° C, R=-1.
- Fig. 12 SEM images of fracture surface of low cycle fatigue samples for Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy at 850 °C with total strain amplitude of: a. $\Delta\epsilon_t/2=\pm 0.32\%$. b. $\Delta\epsilon_t/2=\pm 0.40\%$. c. $\Delta\epsilon_t/2=\pm 0.70\%$.
- Fig. 13 Crack propagation near fracture surface of low cycle fatigue samples for Ti-45Al-8.5Nb-0.2W-0.2B-0.02Y alloy at 850 °C with total strain amplitude of: a. $\Delta\epsilon_t/2=\pm 0.40\%$. b. $\Delta\epsilon_t/2=\pm 0.70\%$.

a

b

C

d

a

b

c