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Investigation on low cycle fatigue behaviors of the [001] and [011] oriental single crystal superalloy at 760 °C

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ABSTRACT

Low cycle fatigue behaviors of a 3Re Ni-base single crystal superalloy with the [001] and [011] orientations have been investigated. At 760 °C, both the [001] and [011] specimens fracture along crystallographic planes. However, the results also indicate a large difference in fatigue deformation behaviors of these two oriented specimens, which is reflected in different deformation mechanisms. For the cyclic stability behavior of the [001] specimens, uneven cross-slip and tangled dislocations and continuous stacking faults cause local stress concentration in the matrix and have an influence on work hardening, while the stress concentration is reduced by dislocation pairs and stacking faults shearing γ' phases. With respect to the [011] specimens, high density matrix dislocations, parallel dislocations and secondary cracks increase resistance to dislocation movement, resulting in cyclic hardening. We discovered continuous stacking faults that have been rarely reported in the low cycle fatigue. Also, these findings might bring about substantial insights that fatigue fracture and deformation mechanisms are responsible for low cycle fatigue anisotropy at intermediate temperature.

1. Introduction

Ni-base superalloys are widely applied as turbine blades and vanes in the aircraft engines and industries gas engines due to excellent physical and mechanical properties [1]. The blades are subjected to complex service conditions, especially in the start-up and shut-down procedures [2–5]. Low cycle fatigue (LCF) can approach real conditions of the blade roots during actual service and evaluate fatigue damage effectively. Therefore, investigating the LCF behaviors is significant for turbine blades design to satisfy the security and dependability requirements.

Although blades are designed to place main stress direction along the [001] crystal orientation due to its great mechanical property, a few degrees deviated off the ideal orientations is inevitable in the blades preparation process. Moreover, blade shape is complicated and multiaxial stress is subjected on the root of blades, generating complicated air-cooling passages and high temperature gradient, which may lead to localized multi-directional stress and complex thermal stress in the blades under service [5,6]. These factors may cause the actual loading direction to be not ideal [001] direction. As is known to all, mechanical properties of single crystal superalloys are remarkably anisotropic. Mackay et al. [7,8] divided stress rupture lives for Mar-M247 and Mar-M200 single crystals into different regions at 760 °C, and pointed that mechanical properties are extraordinarily sensitive to crystal orientations. The shortest stress rupture lives is exhibited by crystals near the [011] orientation, but the stress rupture live is longer when the orientation closed to the [001 − 011] boundary of the stereographic triangle than to the [001 − 111] boundary. However, with regard to the orientation dependence of the LCF, the [001] oriented alloys have the longest life while the [111] oriented alloys have the shortest life [9,10]. Even though the elastic modulus is of great importance, there are also other factors influencing the fatigue life [11,12].

With respect to deformation behaviors and deformation mechanisms during the LCF tests, there have been few published reports, especially for the alloy with other orientations rather than the [001] orientation. Some studies show that the deformation behavior is mainly determined by temperature, which is manifested by cyclic softening at high temperature and cyclic hardening at low and intermediate temperatures [13–15]. Conversely, it does not seem to be applicable for the fatigue deformation of Re-containing superalloys. Li et al. [16] found that the [001] and [011] oriented DD6 alloy containing 2Re exhibit cyclic hardening at 980 °C, while the deformation mechanisms of each oriented alloy was not analyzed. Subsequently, our previous research [17] clarified that the oriented specimens are mainly characterized by

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cyclic softening at the early stage at 980 °C, owing to the formation of dislocation networks and γ' degradation. Specially, under small cyclic amplitude, parallel aligned dislocations in the [011] specimens cause initial cyclic hardening. On the basis of these results, differences in dislocation movement were also considered causing different fatigue deformation behaviors of oriented alloys. Additionally, comparing fatigue properties of Re-containing alloys and Re-free alloys, Li et. al [18] found that the difference in cyclic number and deformation characteristics of this two alloys is large at intermediate temperature, but this research is only limited to the [001] oriented alloys. Therefore, it is necessary to explore fatigue behaviors and deformation mechanisms of Re-containing alloys with other orientations at intermediate temperature, contributing for the blade design and life prediction.

In this work, the low cycle fatigue behaviors at 760 °C of a single crystal superalloy containing 3Re has been systematically investigated. The fatigue fracture modes for the [001] and [011] specimens have been clarified. Additionally, the microscopic deformation mechanisms for the [001] and [011] specimens have been proposed in detail, and the effects of crystal orientations on the deformation behavior in LCF have been discussed.

2. Experimental procedures

The nominal chemical composition of a Re-bearing Ni-base single crystal superalloy used in this study was shown in Table 1. The single crystal rods with directions close to < 001 > and < 011 > crystal orientations were grown on a pre-fabricated seed of the desired orientation and produced by a directional solidification technique. Then the crystal orientation of the cast rods was determined by electron backscattered diffraction (EBSD) technique, bars only within 10° deviations from normal orientations were selected for the LCF tests. All specimens received the heat treatment as follows: 1300 °C × 2 h air cooling + 1120 °C × 4 h air cooling + 1080 °C × 4 h air cooling + 900 °C × 4 h air cooling.

The cylindrical specimens were machined into LCF specimens which had a gauge length of 16 mm and a diameter of 6.5 mm. After machining, the gauge section and excessive arc segment were polished with 2000# sandpaper at a low speed and then subjected to mechanical polishing before fatigue tests to prevent premature crack initiation at the surface machined scratches. The LCF tests were performed under controlled total strain at 760 °C using a servo hydraulic testing machine. A triangular waveform with a constant strain rate of $5 \times 10^{-3}\ s^{-1}$ and a strain ratio of $R = -1$ were used. After fatigue tests, the surface morphologies and the slip patterns on surface were observed by a JSM-7100F field emission scanning electron microscope (SEM). In the following, microstructural observations of the specimens after fracture were carried out by a JEOL JEM-2100 transmission electron microscope (TEM) operating at 200 kV.

3. Results and discussion

3.1. Fatigue life

The relationship between total strain ranges and the fatigue life at 760 ℃ is plotted in Fig. 1a, where the fatigue life of the [001] specimens is longer than that of the [011] specimens. The difference in LCF life between two oriented specimens is much smaller under stress ranges in Fig. 1b. The above results are similar to earlier reports of LCF, the difference of the LCF life is mainly attributed to the modulus of elasticity [9,10]. The elastic modulus of the [001] specimens (97 GPa) is lower than that of the [011] specimens (184 GPa). From the curve of inelastic strain ranges vs. fatigue life (Fig. 1c), an interesting discovery that the [001] specimens also have a longer fatigue life under the condition of inelastic strain ranges. Inversely, there are almost no orientation dependence on N4 [9] and Mar-M200 [10] alloys under inelastic strain ranges, while our experimental alloy manifests the fatigue life of the [011] specimens longer than that of the [001] specimens at 980 ℃ [12]. Based on the above results, it can speculate that the orientation dependence of fatigue life may also be associated with different fatigue failure and deformation mechanisms.

Table 1
Nominal composition of the alloy (wt%).

Elements	C	Cr	Mo	Co	W	Ta	Al	Hf	Re	Ni
Composition	0.07	7	1.5	7.8	5	6.6	6	0.15	3.0	Bal.

3.2. Cyclic stress response behavior

Fig. 2 presents cyclic stress response curves of the [001] and [011] specimens at 760 °C under various strain ranges, which displays different deformation characteristics between two oriented specimens. Cyclic stability is the main feature for the [001] specimens, except for the specimen under the largest strain range which shows cyclic hardening. On the other hand, curves of the [011] specimens show obvious cyclic hardening under most of strain ranges. Notably, deformation behaviors of the experimental alloy are contradictory to that of PWA1480 alloy, the latter with the [001] and [011] orientations are characterized by cyclic hardening at 650 °C [15]. Besides, the opposite results of deformation behaviors between experimental [001] and [011] specimens at 760 °C may be related to deformation mechanisms, which will be analyzed and discussed in detail in the following sections.

3.3. Surface morphologies

The SEM observations of fracture surfaces of experimental specimens with [001] orientation after failure at 760 °C are illustrated in Fig. 3. The most evident feature is that cracks propagate along several intersecting conjugate slip planes, resulting in the fracture surface composed of different smooth facets macroscopically (Fig. 3a). The direction of crack propagation is normal to the intersection line of {111} slip planes, which is clearly visible on these smooth fracture planes indicated in Fig. 3b. It means that crystallographic fracture mode dominates the fatigue fracture of the [001] specimens under this condition. Moreover, multiple crack initiations are originated from the surface of the specimens, as shown in Fig. 3c.

Typical fracture surfaces of the [011] specimens after failure at 760 °C are shown in Fig. 4, which are much rougher than those of the [001] specimens macroscopically. Lots of secondary cracks are formed during the process of crack growth, as a result of interaction among active octahedral slip systems. The secondary cracks as well as fracture surfaces are marked in Fig. 4b according to octahedral slip planes illustrated in Fig. 4c. Consequently, it is believed that the [011] specimens also fracture along crystallographic planes.

Fig. 5 summarizes surface morphologies near fracture surfaces of the [001] and [011] specimens at 760 °C. For the [001] specimens, numerous slip bands with different directions appear on the surface, and each of which is nearly 45° inclined to the stress axis (Fig. 5a). When the crack is long enough, an obvious zigzag feature of the crack propagation is formed by the crack growing along two sets of intersecting slip planes (Fig. 5b). Finally, the crack turns into one dominant slip plane propagating until failure. Compared to the [001] specimens, slip bands with different directions distribute more concentrated and closer to fracture surfaces of the [011] specimens, which are only observed in high magnification Fig. 5c and d. Moreover, it can also be seen that slip bands in the [011] specimens are much thinner and distribute more homogeneously than those in the [001] specimens during the LCF.

3.4. Dislocation movement and deformation mechanisms

Fig. 6 illustrates typical deformation microstructures of the [001]