In-situ observations of the effects of orientation and carbide on low cycle fatigue crack propagation in a single crystal superalloy

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Received 7 March 2010; revised 12 March 2010; accepted 15 March 2010

Abstract

Based on in situ observations with scanning electron microscopy (SEM), we investigated the low cycle fatigue (LCF) propagation behaviors of a Ta-rich single crystal nickel-based superalloy at 550°C in vacuum. Specimens in three different orientations, i.e. [010], [011] and [111], were tested in stress-controlled tension-tension LCF tests with R=0.1. Slip traces can be observed ahead of the crack and distinct crystallographic fracture was dominated. By analysis on the surface slip traces and the crack propagation planes, the operated slip systems were examined to be octahedral slip and the active octahedral plane was identified in this context. The whole cracking process of small fatigue cracks were meticulously tracked and recorded. The evaluation of short crack growth rate for specimens in three orientations was carried out by fracture mechanics approach. The influence of crystallographic orientations on fatigue crack growth are analyzed and discussed. Especially, the interaction between the short crack and local microstructure such as the carbides ahead of the crack were analyzed, altering the propagation path as well as influencing the crack growth rate. The parameter of crack opening displacement was examined to correlate well the crack growth data in different orientations.

Keywords: single crystal superalloy; low cycle fatigue; crack propagation; slip

1. Introduction

Single crystalline (SX) nickel-based superalloys have been more and more widely used as high temperature materials in aircraft turbines as well as in land-based gas turbines [1]. For larger SX turbine blades required in the latter, which contains larger defects such as casting micropores or freckles, short cracks can be easily initiated therein and lead to fatigue failure [1, 2]. For low cycle fatigue concerned regarding to the cyclic plastic deformation during start up, shut down or maintenance, understanding of the short crack propagation can be of significance in the fatigue life prediction and material design of turbine components. It’s known for polycrystalline alloys that the short
crack can propagate much more rapid than the long crack and what’s more, it shows great dependence on the local microstructures, such as grain boundaries and inclusions etc [3]. For precipitate-hardened SX superalloy without grain boundaries, experimental studies on short/long fatigue crack propagation of different alloy systems haven’t produced conclusive results [1, 3-5].

Moreover, the anisotropy of SX superalloy makes it more difficult the evaluation of fatigue crack growth, due to its distinct dependence on crystallographic orientation [6]. Different slip systems may be operated under differing crystallographic orientations and test conditions [4, 6]. The fatigue failure on specified slip planes is known governed by the resolved shear stress, instead of the maximum principle stresses in conventional polycrystals. Hence, fatigue criteria for polycrystalline metals should be validated before application in modeling SX superalloys. So far, the crystallographic fracture mode is still not well accounted for in the current life prediction systems for single crystals. One promising parameter proposed by Chen and Liu [7] based on fracture mechanics is the resolved shear stress intensity AK \_\_ on all 12 {111}<110> slip systems present in Ni-based SX superalloys. While the AK \_\_ parameter has succeeded in modeling octahedral failure mode, it was found to fail in collapsing the fatigue crack growth data of PWA1480 in some single crystal orientations [5]. To improve the octahedral driving force parameter, Telesman and Ghosn [5] proposed a composite parameter AK \_\_ by the geometric average of AK \_\_ and AK \_\_. On the other hand, Wu et al. [8] developed a model based on the continuously distributed dislocation theory (CDDT) and the Stroh formalism to evaluate the fatigue crack propagation in anisotropic materials by crack-tip opening displacement (CTOD). This parameter showed promise in correlating the crack growth rate of Udimet 720 single-edge notched-bend specimen.

In the present paper, the fatigue cracking behaviors in [010] [011] and [111] single crystal superalloy at 550°C were investigated by in situ SEM. The observed results were analyzed by identifying the operated slip systems responsible for growing the crack. The influence of carbides on fatigue short crack growth was examined by real time observation. Small crack growth with respect to crystal orientations were evaluated and discussed accordingly.

2. Introduction

2.1. Structure

The material used in this study was a single crystal nickel-based superalloy, which was developed for fabricating high-performance gas-turbine blades [2]. The nominal composition of this SX alloy is (in wt.%): 0.067 C, 3.9W, 12.0 Cr, 9.0 Co, 3.6 Al, 5.0 Ta and 60.0 Ni etc. The slab specimen had a dog-bone shape with a 2.5mm by 0.4 mm gauge cross section (Fig. 1). The specimens were fabricated from the single bars using spark cutting. The U-shape notch on the edge of each specimen was introduced by a blade. Specimens in three representative orientations were used within this experimental extent, i.e. [010] [100], [011] [100] and [111] [01T] (Figs. 2(a)-(c) respectively). That is to say, the specimen/loading axis is along [010], [011] and [111] respectively, with the transverse orientation (crack growth direction) in [100], [100] and [01T] correspondingly. The notch orientation and loading axis are defined as X and Y axis, respectively. The Z axis is taken to be normal to the X-Y plane, as shown in Fig. 2. Take the specimen in [010] orientation for example, it is known to show multiple octahedral slip, i.e. {111}<110>. Fig. 2(a) shows one of the octahedral slip plane, i.e. (TTI), and the corresponding three slip directions, on which the resolved shear stresses are denoted as τ_n, τ_11, τ_0, respectively.

The polished faces of the samples were etched in a aqueous solution of 4g CuSO_4 + 20mL HCl + 20mL H_2O at room temperature prior to fatigue tests in order to reveal the prevailing microstructure for in situ observations under SEM. No effect of the etching on the crack path was found. The fatigue crack propagation tests were performed in the vacuum chamber of the SEM using a specially designed servo-hydraulic testing system. This machine provided pulsating (sine wave) loads at 10 Hz of ±1 kN maximum capacity and a displacement range of ±25 mm. The signal of the SEM was directly transferred to a computer via a direct memory access type A/D converter, making it possible to sample 960×1280 frames of SEM images successively. The SEM was operated at an accelerating voltage of 15 kV.

A constant maximal stress (850 MPa for [010] and [111]) was adopted throughout the test to study the fatigue crack propagation process. For [011] specimen, the maximal stress was lowered from 850MPa to 750MPa after 1000 cycles due to overly fast crack growth. The waveform utilized was sinusoidal. All fatigue tests were load
controlled at a stress ratio of 0.1 with a loading frequency of 2 Hz. Images of fatigue crack with different lengths were taken in situ at different cycles of loading and hence the crack growth rate can be calculated from the measured crack lengths. By comparing with the propagation process, the dependence on the local structures and cracking manner can be analyzed and disclosed.

Fig. 1. Slab specimen in fatigue test (dimensions in mm)

Fig. 2. Specimens in three crystallographic orientations (a) (010) [100]; (b) (011) [100]; (c) (111) [011]

3. Results and discussion

3.1. Features of fatigue crack growth

Fig. 3(a) compares the crack growth of three orientations in vacuum at 550°C in terms of fatigue loading cycles. In general, the curves follow a linear relationship in the log-linear coordinate. This is consistent with Molent’s review on a range of aircraft aluminum alloys [9], following the early work of Frost and Dugdale [10], reported that crack growth under constant amplitude loading could be described via a simple log linear relationship:

\[ a = a_0 e^{\lambda N} \]

where \(a_0\) is the initial flaw size, and \(\lambda\) is a constant. The result here showed that this relation hold in the short crack growth for anisotropic single crystals. Moreover, in this framework (Eq. (1)), the fatigue crack growth rate (FCGR) can be given by
\[ \frac{da}{dN} = \lambda a e^{2n} \]  

(2)

It is seen in Figs. 3 (a)-(b) that Eqs. (1) and (2) can be used to well characterize the small crack growth in SX superalloy for three different orientations. Moreover, the crack growth rate was plotted versus the crack length, as shown in Fig. 3(c). Generally, crack growth rate is enhanced with the increase of crack length due to larger stress intensity. No evident relationship can be found due to scattering, especially for the [111] specimen.

Fig. 3. Fatigue crack growth (a) Crack length as a function of number of cycles; (b) FCGR versus number of cycles; (c) FCGR versus crack length (notch depth not included)

Further, the crack growth rate was evaluated by nominal stress intensity factor \( \Delta K \), i.e. the well known Paris law. It is noticed in Fig. 4 that the FCGR show evident dependence on the crystallographic orientation. Especially, the one in [111] shows much lower FCGR, and great fluctuation is observed during different propagation stages. Those in [010] and [011] show less difference considering the systematic scattering. The FCGR in [011] is slightly higher although the applied stress is a bit lower, which can be attributed to the lower yield stress and tensile strength in [011] compared to the other two orientations.

Fig. 4. Fatigue crack propagation rate in terms of \( \Delta K \)

3.2. In situ observations of crack propagation

The in situ observations during the whole propagation process provide evidence for us to better understand the effect of orientation and microstructure on the fatigue crack growth. Fig. 5 shows the cracking behaviors of [010] specimen. The crack path is about 45° inclined to the loading axis, indicating octahedral cracking mode. This failure mode was also reported in PWA1484 superalloy by Telesman et al [5]. It supports that at intermediate temperatures, crack growth of SX superalloy occurs either on a single octahedral slip plane or on multiple different slip, resulting in large crystallographic \{111\} facets on the fracture surface [2]. Further, crack bifurcations commonly occurred when the inclined crack was blocked by carbides in the [010] specimen, as shown in Figs. 5(b) and (c). The crack
would turn away from the slip system towards Mode-I direction for a few microns and return to the octahedral plane subsequently. Note that Figs. 5(a)-(d) corresponds to A1-A4 in Fig. 4. The effect of carbide on impeding the crack propagation and altering the crack path is evident. An increase in crack growth rate can be noticed after the crack passed through the carbides. Although [010] is known to be multiple slip orientation, slip along single octahedral plane is remarkable in this observation, i.e. (TT1), driven by resolved shear stress as illustrated in Fig. 2(a).

Fig. 5. Fatigue crack growth behavior for [010] oriented specimen (a) 24973 cycles, 30.75 μm; (b) 25980 cycles, 68.17 μm; (c) 26374 cycles, 104.38 μm; (d) 26421 cycles, 133.71 μm

Fig. 6. Fatigue crack growth behavior for [011] oriented specimen (a) 1054 cycles, 63.58 μm (b) 1290 cycles, 72.34 μm (c) 1515 cycles, 159.79 μm (d) 1543 cycles, 185.78 μm
For [011] oriented specimen, under maximal stress of 850MPa, crack growth along single octahedral slip plane was monitored for the first 1000 cycles (similar to Fig. 6(a)), accompanied with a fast growth rate. The maximal stress was then lowered to be 750MPa for the following study of crack growth, revealing obvious duplex slip as indicated by the zigzag cracking mode in Fig. 6(b). The observation here supports that the failure mode is greatly dependent on stress intensity, i.e. single crystallographic fracture mode was favored under high stress intensity, whereas macroscopic Mode-I like fracture morphology formed by two sets of intersecting slip planes under low $\Delta K$ levels. The intersecting slip traces near crack tip is evident in Fig. 6(b)-(c). The operative octahedral slip planes are illustrated in Fig. 2(b). According to crystallographic analysis based on the slip traces and crack propagation direction, the cracking planes are identified to be along $(1\bar{1}1)$ and $(111)$ alternately. According to Telesman et al. [5], the slip systems with high resolved shear stress intensity are preferential to be operated during cyclic loading. With the advance of crack, there is an indication of promoted Stage I crystallographic cracking (see Fig. 6(c)), due to a concomitant increase in $\Delta K$. The crack finally collapsed with a casting pore in the superalloy and resulted in a sharp increase in crack length, leading to the catastrophic rupture (Fig. 6(d)).

![Fatigue crack growth behavior for [111] oriented specimen](image)

Fig. 7. Fatigue crack growth behavior for [111] oriented specimen (a) 36103 cycles, 53 $\mu$m (b) 51694 cycles, 144.9 $\mu$m (c) 57231 cycle, 251.49 $\mu$m (d) 66103 cycle, 386.16 $\mu$m (e) 67699 cycles, 425.52 $\mu$m (f) 67751 cycles, 504.41 $\mu$m

For the [111] specimen, although the (111) plane with a <011> slip direction is in the plane of nominal crack growth, stage I (by shearing) crack growth occurred preferentially along other {111} planes, under pure mode I loading. Single slip system was dominated when the crack length was beyond 20$\mu$m from the notch. Of interest is that a secondary crystallographic crack can still occur parallel to the current cracking plane, and evolve into the main crack as show in Fig. 7(b). The crack path is measured to be about 22° inclined with respect to the loading direction. The dominated slip plane is identified to be the octahedral plane (1\bar{1}1), by searching all possible slip systems either octahedral or cube. Figs. 7(a)-(f) corresponds to C1-C6 in Fig. 4. The great dispersion in FCGR of [111] specimen can be rationalized by the following observations. In Fig. 7(c), impeding effect of carbide on crack growth is obvious and the crack propagation path is altered, corresponding to a decrease in FCGR. The crack which hit a carbide either meandered propagation direction or propagate through it, depending on the local stress concentration and the strength of carbide. Mode-I crack would emerge when the crystallographic cracking along slip plane was blocked, partly because that the inclined cracking is less efficient in increasing $\Delta K$ compared to the transverse crack, as observed in Fig. 5(b). In Figs. 6(d) and (e), the crack tip swerved when impeded by carbide and
the transverse crack finally evolved into the main crack and led to an evident increase in propagation rate. In Fig. 6(f), multiple cracking parallel to the initial octahedral plane is observed and the main crack returns to propagate along it. The crack growth rate is promoted, leading to the final fracture.

It is revealed from above in situ observations that octahedral slip induced cracking is the dominant mode in the present test context. The crack would turn from the octahedral plane towards the Mode-I like direction when blocked by microstructural barriers, e.g. carbides, and return to propagate along the same slip plane afterwards. This phenomenon is evident especially for [001] and [111] specimens. The effect of stress intensity factor on crack growth was disclosed by the [011] specimen. Accompanied with the octahedral cracking, the dominated deformation mechanism is shearing of $\gamma'\gamma$ at 550°C, as we observed in the low-cycle fatigued specimens under intermediate temperatures [2].

### 3.3. Evaluation of fatigue crack growth

For the evaluation of fatigue short crack, some approaches have been successful in correlating the short crack growth of alloys, e.g. crack opening displacement (COD) [1, 11], plastic zone size (PZS) ahead of crack tip [4], effective stress intensity factor [5, 9] and so on. For example, Shyam and Jones et al. [11] evaluated the short crack growth behaviors of east aluminum alloys by COD based on the continuous distributed dislocation theory proposed by Bilby, Cottrell and Swinden (BCS) [12]. The cyclic COD under a far field stress $\sigma_{\text{max}}$ with a stress ratio of $R$ is given by

$$\phi_t = \frac{16\sigma_{\text{max}}(1-\nu^2)a}{\pi E} \ln \left( \sec \left( \frac{\pi\sigma_{\text{max}}(1-R)}{4\sigma_y} \right) \right), \quad (3)$$

where $\nu$ is Poisson’s ratio, $a$ is the crack length, $E$ is Young’s modulus, $\sigma_y$ is the yield strength. This model can take into account the influence of modulus and yield properties on crack growth and is initially used in the formulation for polycrystalline materials. Assuming that a line crack consists of dislocation pile-ups with the resultant stress satisfying the crack surface boundary condition, following the BCS approach, Wu et al. [8] considered the COD of a crack in anisotropic elastic-perfectly-plastic materials under mode I loading,

$$u = \frac{4a\sigma_{\text{max}}}{\pi F_{22}} \ln \left( \sec \left( \frac{\pi\sigma_{\text{max}}}{2\sigma_y} \right) \right), \quad (4)$$

where $F_{22}$ is an element of material’s elastic matrix solved according to Stroh formalism [8]. It is noted that Eq. (4) was derived on the basis of the plane strain crack condition. Nevertheless, Eq. (4) shows a similar dependence of COD on the elastic and yield properties for anisotropic materials, except that Eq. (3) takes into the effect of reverse flow and thus includes the stress ratio $R$. Hence, it seems that for the fatigue crack growth in the current slab specimens, in which plane stress is dominant for most of the time, the COD $\phi_{t,\text{SS}}$ can be assumed to follow

$$\phi_{t,\text{SS}} \approx \phi_t$$

or

$$\phi_{t,\text{SS}} \approx \frac{\sigma_{\text{rel}}(\{hkl\})a}{\pi E_{\{hkl\}}} \ln \left( \sec \left( \frac{\pi\sigma_{\text{max}}(1-R)}{4\sigma_{\text{rel}}(\{hkl\})} \right) \right), \quad (5)$$

In Eq. (5), $\sigma_{\text{rel}}(\{hkl\})$ and $E_{\{hkl\}}$ is the yield strength and modulus in the direction of applied stress, so as to incorporate the dependence on crystal orientation $[h k l]$. To evaluate the parameter proposed in Eq. (5), the crack growth data in three orientations was plotted as a function of this parameter, shown in Fig. 8. It is noticed that this parameters can collapse the data of three orientations into a narrow band. Dispersion was reduced in comparison with Fig. 3. The results obtained here support the validity of using the COD parameter as the characteristic
parameter for octahedral fracture in single crystal superalloys, which was also proposed in Wu’s studies of correlating the FCGR data of Udimet 720 single crystal [5]. Considering the limited samples in this experimental context, more test data is required to check the applicability of this parameter in future works.

Fig. 8. Fatigue crack growth data in terms of COD

4. Conclusions

The fatigue short crack behaviors of a single crystal nickel-based superalloy in three different orientations ([010], [011] and [111]) were studied at 550°C in vacuum by in-situ SEM observations. Crystallographic cracking mode is dominated for all specimens and slip traces can be commonly observed ahead of the crack on specimen surface. Local microstructural inhomogeneities such as the carbides in the front of crack were found to expressly alter the propagation path as well as decrease the crack growth rate. The operated slip systems for [010], [011] and [111] specimens were identified to be octahedral slip, by analysis on the surface slip traces and the crack propagation direction. The crack growth rate for specimens in three orientations was evaluated by fracture mechanics approach. The parameter of crack opening displacement is found to be able to collapse the crack growth data of three orientations into a narrow band the present test.

Acknowledgements

The financial support from the National Natural Science Foundation of China (No. 10872105) is highly acknowledged.

References


