6th International Conference on Creep, Fatigue and Creep-Fatigue Interaction [CF-6]

Subcritical Crack Growth on Crystallographic Planes in a Ni-base Superalloy: Relevance to Orientations

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Abstract

Cast Ni-based superalloys, essential materials for gas turbines, possess large grains comparable to wall thickness of structural components: e.g., blades and vanes. Fatigue cracks propagating in these kinds of materials must be significantly influenced by crystallographic factors of materials: e.g. orientation of grains and grain boundaries. This paper covers such a basic problem. For the purpose, at first an effect of crystallographic orientation on the crack propagation rate is introduced, using a single crystal Ni-based superalloy CMSX-4 specimen with different primary and secondary orientations. Discussions are also made on what mechanics and mechanisms control the fatigue crack propagation behavior, on the basis of anisotropic fracture mechanics.

Keywords: Fatigue crack; single crystal Ni-based superalloys; orientation; mixed mode fracture; slip plane and shear

1. Introduction

High temperature components in advanced industrial gas turbines, such as blades or vanes, have to operate under a severe combination of mechanical and environmental loadings. These loadings inevitably cause many types of damages and degradations in components during the operation period. Materials used in them, especially Ni-base superalloys, are so expensive items that they can be a substantial part of an operating and a maintenance budget [1-7]. Therefore, it is of considerable technical and financial benefit if the condition of the blades can be assessed and their remaining life determined. Whilst some efforts have been made to estimate the remaining life based on the crack assessment, there has not been well-established method and methodologies yet [1-7]. Some reasons are responsible for these difficulties. One of them is from such a current state that little has been known quantitatively about their fatigue properties, those may often significantly depend not only on the primary but also secondary crystallographic orientations [4,5]. The interaction between crack and grain boundaries has not been known well, while it may be significant especially in cast Ni-base superalloys because the dimension of components is generally in same order as grain size of material [9]. In addition, there

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are also technical hurdles to get test results using a miniature samples; here knowledge is required from local level and not from macro level by means of standard size specimens [6].

From these background, it is the final objective of this work to quantitatively make clear the crack propagation behavior interacting with grain boundaries. The present research flow consists of three stages. The first stage is to experimentally make clear the fatigue crack propagation resistances of Ni-base superalloys, using a miniature compact tension (CT) specimens. The second is to investigate the effect of crystallographic orientation on the crack propagation rate in single crystal superalloy specimens with different orientations, based on anisotropic fracture mechanics.

2. Experimental procedures

2.1. Materials

A single crystal superalloy, CMSX-4, is tested in this work. The chemical compositions (wt.%) of CMSX-4 and the condition of heat treatments are summarized in Table 1. The microstructure of CMSX-4 is given in Fig. 1, showing that the size of cuboidal gamma-prime precipitates and the volume fractions are approximately 0.5 μm, and 63 %, respectively.

![Fig. 1. Microstructure of CMSX-4.](image)

Table 1. Chemical compositions (wt. %) of CMSX-4 and condition of heat treatment.

<table>
<thead>
<tr>
<th>Al</th>
<th>Ti</th>
<th>Ta</th>
<th>Mo</th>
<th>W</th>
<th>Re</th>
<th>Hf</th>
<th>Cr</th>
<th>Co</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>5.7</td>
<td>1.0</td>
<td>6.5</td>
<td>0.6</td>
<td>6.4</td>
<td>2.9</td>
<td>0.1</td>
<td>6.4</td>
<td>9.7</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

Solution treatment (by 1277°C×2h + 1296°C×3h + 1304°C×3h + 1313°C×3h + 1316°C×3h + 1321°C×3h + 1321°C×3h), followed by aging (by 1080°C×4h + 880°C×20h).

2.2. Specimen preparation

One of purposes of this work is to quantitatively evaluate the fatigue crack propagation resistance of superalloys for hot section components of gas turbines. Here it is important to remind that the thickness of hot section component is generally in order of milli-meters. Thus, a miniature specimen must be suitable for this work. In this work the miniature compact tension (CT) specimens illustrated in Fig. 2 were extracted from the cast superalloy rods for the fatigue test [6]. The geometry of specimen was designed so that it was almost proportional to that recommended in ASTM E647 [8], except for the specimen thickness. The experimental variables in the CMSX-4 specimen are both the primary crystallographic orientation of loading axis (that is, [100], [110] and [111] directions) and the secondary orientations, as briefly summarized in Table 2. These orientations are not exact but approximately within a mismatch of 5 degree. Hereinafter the respective specimens are expressed with the designations in Table 2. In the IN939, on the other hand, the miniature
compact tension samples were extracted from the center of rod bar so that the specimens belong to so-call RL orientation.

2.3. Fatigue crack propagation tests

After all the specimen surfaces were polished to a mirror-like finish using emery papers and Al\textsubscript{2}O\textsubscript{3} powder, fatigue crack propagation tests were carried out using a servo-electro hydraulic machine at room temperature. Here, a loading frequency and a loading ratio $R (=K_{\text{min}}/K_{\text{max}})$ were 10 Hz and 0.4, respectively. The crack length was monitored by means of a traveling microscope with a resolution of 0.01 mm from both the specimen surfaces.

![Fig. 2. Geometry of miniature CT specimen used.](image)

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Orientation of loading axis, L</th>
<th>Orientation of secondary axis, T</th>
<th>Orientation P</th>
<th>Apparent growing direction, B</th>
<th>Inclined angle, $\theta$ (deg.)</th>
<th>Inclined angle, $\pi$ (deg.)</th>
<th>Dominant fracture surface plane</th>
</tr>
</thead>
<tbody>
<tr>
<td>100-1</td>
<td>[0 0 1]</td>
<td>[1 0 0]</td>
<td>[0 1 0]</td>
<td>[0 1 1]</td>
<td>45</td>
<td>45</td>
<td>[1 1 1]</td>
</tr>
<tr>
<td>100-2</td>
<td>[0 0 1]</td>
<td>[1 0 0]</td>
<td>[0 1 0]</td>
<td>[0 1 1]</td>
<td>45</td>
<td>45</td>
<td>[1 1 1]</td>
</tr>
<tr>
<td>110-1</td>
<td>[2 2 1]</td>
<td>[1 1 0]</td>
<td>[1 1 4]</td>
<td>[1 1 2]</td>
<td>55</td>
<td>0</td>
<td>[1 1 1]</td>
</tr>
<tr>
<td>110-2</td>
<td>[1 1 0]</td>
<td>[1 1 2]</td>
<td>[1 1 1]</td>
<td>[1 3 2]</td>
<td>22</td>
<td>58</td>
<td>[1 1 1]</td>
</tr>
<tr>
<td>110-3</td>
<td>[1 1 0]</td>
<td>[0 0 1]</td>
<td>[1 1 0]</td>
<td>[1 1 0]</td>
<td>0</td>
<td>35</td>
<td>[1 1 1]</td>
</tr>
<tr>
<td>111-1</td>
<td>[1 1 1]</td>
<td>[1 1 2]</td>
<td>[1 1 0]</td>
<td>[3 1 2]</td>
<td>68</td>
<td>55</td>
<td>[1 1 1]</td>
</tr>
</tbody>
</table>

The apparent crack propagation rate, $da/dN$, was evaluated in terms of crack length, $a$, that is a projected length to normal line to loading axis, conforming to ASTM standard, E647 [8]. In order to minimize the effect of initial EDM notch, the crack growth behavior was analyzed after the crack length from the initial EDM
notch root was longer than 0.3 mm. The stress intensity factor range, $\Delta K_{app}(\alpha)$ was calculated in terms of $\alpha$, conforming to ASTM standard, E647 [8];

$$\Delta K_{app}(\alpha) = F_i(\xi)\sigma_\infty\sqrt{\pi a}$$

with

$$\sigma_\infty = \frac{6\Delta P}{BW^2}, \quad \xi = a/W$$

$$F_i(\xi) = \frac{1}{6\xi^{3/2}} (29.6 - 185.5\xi + 656\xi^2 - 1017\xi^3 + 639\xi^4)$$

where $\Delta P$, $B$, $W$ and $\sigma_\infty$ are load range, specimen thickness, specimen width, and apparent remote nominal stress at crack tip, respectively. However, as noted later, the above treatment is no more than conventional, because the crack propagated on planes significantly deviating from macroscopic principal planes in this work.

3. Results and discussion

3.1. Apparent effect of crystallographic orientation on fatigue crack growth rate in single crystal alloy

Figure 3 reveals the relationship between the crack propagation rate, $(da/dN)_{app}$, and the stress intensity factor range, $\Delta K_{app}$, in the respective specimens, where $(da/dN)_{app}$ and $\Delta K_{app}$ are evaluated based on the incremental amount of projected crack length, $D_a$, normal to loading axis on the specimen surface. Here $\Delta K_{app}$ was evaluated by Eq. (1). This type of treatment follows a method employing fracture mechanics [8]. It is found that the crack propagation rates significantly depended on the specimen orientation, or loading axis direction. Generally speaking, a series of the [100] specimens revealed the highest, and the [111] specimens did the minimum, respectively. On an appearance, this order corresponds to the orientation dependence in anisotropic elastic modulus to the loading axis in the respective specimens [4]. However, it should be noted that the above dependencies are no more than an apparent tendency, because the treatment is based on projected crack length normal to loading axis observed on the specimen surface without taking account of intrinsic fracture mechanism and mixed mode fracture. This will be discussed through this work.

![Fig. 3. Relationship between the crack propagation rate, $(da/dN)_{app}$, and the conventional stress intensity factor range, $\Delta K_{app}$](image-url)
Morphologies of some typical fracture surfaces are given in Fig. 4. It is found from Fig. 4 and Table 2 that the fracture surface was significantly inclined to the loading axis, and revealed crystallographic features associating with geometrically high asperities. According to the fractographic analysis by means of scanning electron microscope, the fracture surface was almost composed of \{111\} crystallographic planes. From a macroscopic level the geometrical features of the primary fracture plane was analyzed and summarized in Table 2, where the features are represented by the inclined angles, $\theta$ and $\eta$; see an illustration in Table 2. Note that the fracture surface did not consist of a single plane but multiple planes as seen from Fig. 4; hence, the angles, $\theta$ and $\eta$, indicate the measurement values of the most dominant fracture surface. From a microscopic level, the fractographic analysis suggested that the fatigue crack propagation is associated with slip plane decohesion mechanism; or a shearing dominating mode. These observations indicate that an apparent engineering treatment given in Fig. 3 is not adequate: thus it is essential to introduce mixed mode fracture treatments for an in-depth analysis.

3.2. Intrinsic effect of crystallographic orientation on fatigue crack growth rate in single crystal alloy

Roughly speaking, there must be two kinds of approaches to deal with mixed mode fracture: one is a theory that aims to predict not only a crack extension but also crack advancing plane. So-called circumferential stress maximum criterion [10] and energy release rate maximum criterion [11] are classified into this type of treatment. The other is a macroscopic treatment based on total energy release rate with crack advancing [4,11]. In the latter case, total energy release rate, $G$, is a parameter often used. Here, $G$ is defined by,

$$G \equiv G_I + G_{II} + G_{III} \quad (\text{denoted by } G_{I+II+III})$$

(2)

where $G_I$, $G_{II}$ and $G_{III}$ are energy release rates corresponding to Mode-I, II, III stress intensity factor components. While the $G$ has a limited ability to predict how cracks grow under mixed mode condition, it is a very useful engineering parameter for the present work, because the crack propagation plane was confined on \{111\} crystallographic plane in this work (that is, it is not always necessary to explicitly predict to which direction crack grow). Accordingly, some discussions will be made in terms of $G$ in this section.

General equations of $G_I$, $G_{II}$ and $G_{III}$ has been derived for cubic structure material by considering their elastic anisotropy in ref. [4], those are functions of stress intensity factor ranges. Although it is to be noted that the stress intensity factors are given in terms of projected crack length, $a$, in some cases but in terms of actual crack length, $a'$, in other cases, it is possible to convert between the two cases. Taking account of a geometrical correlation between $a$ and $a'$ referring to an illustration in Table 2, $G_I$, $G_{II}$ and $G_{III}$ for the present case where fracture took place on \{111\} planes can be nearly estimated as follows:
\[
G_l = \frac{(1 - \nu_{111}^2)}{E_{111}} \Delta \sqrt{\pi a'} = \frac{(1 - \nu_{111}^2)}{E_{111}} \Delta K_{app} (a')^2 \frac{a'}{a} (l)^4
\]

\[
G_{II} + G_{III} = \frac{1}{2E_{111}} \Delta K_{app} (a')^2 \left( \frac{F_{II} (a')}{F_{I} (a)} \right)^2 \left( \frac{\Delta \pi_{II}^2 + \Delta \pi_{III}^2}{\Delta \sigma^2} \right)
\]

with

\[
\Delta K_n (a') = F_I (a') \Delta \sigma \sqrt{\pi a'} = \Delta K_{app} (a') \sqrt{\pi a'} / a (l)^2
\]

\[
\Delta K_t (a') = F_{II} (a') \Delta \tau \sqrt{\pi a'}
\]

\[
F_{II} (a') = F_{III} (a') = \sqrt{\frac{2}{\pi}} \tan \left( \frac{\pi \xi}{2} \right)
\]

\[
\left( \frac{\Delta \sigma_n}{\Delta \sigma} \right)^2 = \xi^4
\]

\[
\left( \frac{\Delta \pi_{II}^2 + \Delta \pi_{III}^2}{\Delta \sigma^2} \right) = \left( \frac{\Delta \pi_{II}^2}{\Delta \sigma^2} \right) = \xi^2 (1 - \xi^2)
\]

where,

- \(a'\): actual length of inclined crack,
- \(E_{111}\): elastic constant normal to \{111\} plane,
- \(G_{111}\): shear modulus to shearing direction on \{111\} plane,
- \(\Delta K_{app}\): apparent stress intensity factor range by Eq. (1),
- \(F_I (a')\): boundary correction factor for Mode I loading as function of \(a'\),
- \(F_{II} (a')\): boundary correction factor for Mode III loading as function of \(a'\),
- \(l\): directional cosine between \{111\} plane and loading axis,
- \(\Delta K_n (a')\): stress intensity corresponding to normal stress component to crack plane, with actual crack length, \(a'\).
- \(\Delta K_t (a')\): resolved shear stress intensity, with actual crack length, \(a'\).

Note \(\Delta K_t (a')\) and \(\Delta K_n (a')\) are expressed in terms of remote resolved normal and shearing stresses around crack tip (those are denoted by \(\Delta \sigma_n\) and \(\Delta \tau\), respectively), and their expressions by Eqs. (4) are not explicit but approximated that may have a few percent error to their explicit values. \(F_{II} (a')\) from ref. 12. [12] On the other hand, as shown in the previous section, the fatigue fracture surface was significantly inclined against a normal plane to loading axis in this work. Corresponding to this phenomenon, it is also preferable to convert the fatigue crack growth rates \(\frac{da}{dN}\) that were conventionally evaluated by projected crack length into those on fracture plane, \(\frac{dA}{dN}\). While this type of strict conversion is impossible so far as the crack plane is explicitly expressed, it is possible to estimate the most plausible value; denoted by \(\frac{dA}{dN}\), which must be given by,

\[
\Delta A = \frac{\Delta a}{\cos \theta \cos \eta}
\]
significantly beyond the extent of error. It is also interesting to note that the crack propagation resistance of the [111] specimen do not reveal the highest based on $\Delta K_{\tau}$, a different trend from that in Fig. 3. These strongly suggest that the crack propagation was not primarily driven by opening mode in this work. When they are correlated with $\Delta K_{\tau}$, on the other hand, the dependence of primary crystallographic orientation seems to be little, or minimized in this work: see Fig. 6. This means that the fatigue crack growth behaviors as well as their rates were controlled by shear mode in this work, since $\Delta K_{\tau}$ is an equivalent parameter to represent stress intensity for shear mode fracture. This finding agrees with the fractographic observation, showing that the fracture was taken place by slip plane decohesion mechanism. The findings by other researchers [7] also support this conclusion, which has pointed out the importance of resolved shear stress intensity. Thus, it can be concluded that the crystallographic orientation effect on the crack growth rate may be essentially apparent, and it may be attributed to the difference in shear component to fracture depending on the specimen orientation.

4. Conclusions

The fatigue crack propagation of Ni-based superalloy which possessed large grains comparable to the specimen thickness was investigated at room temperature. Special focus was put on the effects of crystallographic orientation on the growth rates and the interactions with grain boundaries.

The conclusions derived are summarized as follows:

- In single crystal alloy, CMSX-4, the fatigue crack propagation rates significantly depended on the crystallographic specimen orientation. On the basis of apparent Mode I stress intensity factor range, a series of the [100] specimens revealed the highest rates, and the [111] specimens did the lowest. Here, the fatigue fracture was driven on the crystallographic {111} slip planes in all specimens.
- When the crack growth rates in the CMSX-4 specimens with different orientations were correlated with $\Delta K_{\tau}$, a parameter to represent stress intensity to shear mode fracture, the dependence of primary crystallographic orientation almost disappeared. This means that the effect of crystallographic orientation shown in Conclusion (1) was apparent, but the rates may be intrinsically controlled by magnitude of stress intensity to shear mode fracture.
Acknowledgement

Financial supports by the Grain-in-Aid for Scientific Research by JSPS (No. 21246022) is greatly acknowledged.

References