Anisotropic Stress Rupture Properties of a 3rd-Generation Nickel-Based Single-Crystal Superalloy at 1100 °C/150 MPa

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Abstract

The influence of orientation on the stress rupture behaviors of a 3rd-generation nickel-based single-crystal superalloy was investigated at 1100 °C/150 MPa. It is found that the stress rupture anisotropy is shown at 1100 °C, but not so obvious compared with that at intermediate temperatures. The [001] specimens display the longest rupture life, [111] specimens show the shortest rupture life, and [011] specimens exhibit the intermediate life. Detailed observations show that the final fracture is caused by crack initiation and propagation, and the anisotropy of three oriented specimens is related to the fracture modes, γ/γ' microstructures, interfacial dislocation networks and cutting mechanisms in γ' phase. For [001] specimens, *N*-type rafted structures are formed which can well hinder the slip and climb of dislocations. Besides, the regular interfacial dislocation networks can prevent dislocations from cutting into γ' phase, leading to the improvement of the creep resistance. For [011] specimens, $\pm 45^\circ$ rafted structures and irregular networks result in less strain hardening. For [111] specimens, a large number of crack propagation paths and inhomogeneous deformations caused by irregular rafted structures deteriorate the property and result in the shortest life. Furthermore, *a*[100] superdislocations with low mobility are widely formed in [001] and [011] specimens which suggests the low creep strain rate during steady creep stage, whereas superdislocations in [111] specimens possess high mobility, which indicates the high strain rate and corresponding poor stress rupture property.

Keywords Single-crystal superalloy · Stress rupture · Anisotropy · Microstructure · Dislocation

1 Introduction

Nickel-based single-crystal (SX) superalloys are widely used for advanced aero and gas turbine blades due to their excellent creep resistance at elevated temperatures [1–4]. It is well known that modern turbine blades are mainly fabricated along [001] orientation due to its excellent comprehensive properties and the preferred growth direction. However, the SX superalloy is inherently anisotropic and turbine blades are usually manufactured with complicated shapes and cooling channels, which may induce the multi-axial and multi-thermal stresses for blades in service [5]. Moreover, the root

and lower portions of the turbine blades may undergo significant stresses in directions transverse to the major stress axis of the blades [6]. The cracks on the blade platform may be visible, and the propagation of these cracks is affected by the properties in different crystallographic directions [7]. The crystallographic orientation has great effects on creep behavior and creep life of SX superalloys. Therefore, the anisotropic creep properties of SX superalloys are worth of investigation, especially for the specimens with the typical [001], [011] and [111] orientations.

At temperature range of 750–871 °C, the creep behaviors for the specimens with [001], [011] and [111] orientations have been fully investigated, which are thought to be highly anisotropic. Many studies [6, 8–11] indicated that the creep anisotropy at this temperature range was mainly attributed to the activation of different $\langle 112 \rangle \{111\}$ and $\langle 110 \rangle \{111\}$ slip systems for different oriented specimens. With the development of advanced aircraft engines, new generations of SX nickel-based superalloys operate at higher temperatures over 1000 °C. High-temperature deformation of SX superalloys is dominated by several prominent features, such

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as directional coarsening of γ' precipitates [12–14], formation of dislocation networks at γ/γ' interface [15, 16] and γ' phase cutting by $a<110>$ or $a<100>$ superdislocations [17–21], but most of the studies focus on the deformation in [001]-oriented SX superalloys. The creep deformations of [011] and [111] specimens have so far received sporadic attention [9–11, 22–27].

Sass et al. [9, 10] showed that the rafted γ' phase reduced the creep anisotropy of [001]- and [011]-oriented CMSX-4 alloys at 980 °C. No directional coarsening of γ' phase was observed in [111] specimens, which resulted in the poorest life of [111] specimens. Liu et al. [11] thought that variation of rafts could strengthen the [111] specimens and N -type rafts might lead to shorter life of [001] specimens during creep at 1010 °C/248 MPa. However, Su et al. [22] confirmed that N -type rafted γ' phase in [001] specimens could effectively hinder the movement of matrix dislocations and result in the strain hardening. But rafts with 45° and 40°–60° inclined to the loading directions in [011] and [111] specimens, respectively, lowered the creep resistance [22, 23], which are not in line with the results in Ref. [11]. Besides, many studies have shown that γ' phase cutting mechanisms strongly affect the creep behaviors of [001]-oriented SX superalloys [17–19], but less information is available on superdislocations in [011] and [111] specimens. Jácome et al. [24, 25] reported that the lack of $a<100>$ superdislocations with low mobility might be partly responsible for the higher strain rate and poorer life of [011] specimens as compared to [001] specimens, while the configurations of superdislocation in [111] specimens have rarely been reported. Moreover, the final fracture of the SX superalloys during high-temperature creep is mainly caused by creep cavitation damage [28]. Mälzer et al. [27] demonstrated that the aligned pores associated with primary dendrites could affect the final fracture of different oriented specimens, but they were not directly observed.

Therefore, in this paper, the stress rupture anisotropy ([001], [011] and [111] orientations) of a 3rd-generation SX superalloy was investigated at 1100 °C/150 MPa. Main purpose is to clarify the effects of fracture mechanisms, γ/γ' microstructures and underlying deformation mechanisms on stress rupture anisotropy.

2 Experimental

A 3rd-generation nickel-based SX superalloy DD33 with 4 wt% Re was used in the present experiment. The master alloys were melted by vacuum induction furnace and directionally solidified into SX slabs (20 mm × 80 mm × 220 mm) using selector technique in [001] orientation and seeding techniques in [011] and [111] orientations. The crystal orientation of the slabs was determined by electron

backscattered diffraction (EBSD) technique. Only slabs deviating from their precise directions within 10° were adopted. Solution and two-step aging heat treatments were given to the SX slabs as follows: 1330 °C/10 h, air cooling (AC) + 1150 °C/4 h, AC + 870 °C/24 h, AC. After heat treatment, the γ channel widths of different oriented specimens were measured along the cross sections. For [011] specimens, only (100) channel (so called gable channel) width was measured which was the real γ channel width. For [111] specimens, γ channel width was calculated as $a\sin 54.7^\circ$. The parameter a is the measured channel width on the (111) cross section, and 54.7° is the angle between (111) and (001) planes.

Specimens with a gauge length of 25 mm and a diameter of 5 mm were machined from the SX slabs along three orientations. The stress rupture tests were carried out at 1100 °C/150 MPa. Some of the tests were interrupted at 30 h to observe the γ/γ' microstructures. The microstructures and fracture morphologies were observed by optical microscopy (OM) and scanning electron microscopy (SEM). These samples were mechanically polished and etched in a solution of 20 g CuSO₄ + 50 ml HCl + 100 ml H₂O. The surface morphology of [111] specimens was observed by stereomicroscope after being etched in the mixture of hydrochloric acid (40 ml) and hydrogen peroxide (40 ml). Foils for transmission electron microscopy (TEM) were cut from the ruptured specimens, perpendicular or parallel to the applied stress. The foils were mechanically ground to ~50 µm and then electrochemically polished in a solution of 10% perchloric acid and 90% alcohol at –20 °C and 20 V. The dislocations configurations were investigated by JEM-2100, operating at 200 kV.

3 Results

3.1 Microstructures After Heat Treatment

Figure 1 shows the heat-treated microstructures of γ/γ' phases in [001], [011] and [111] specimens. It is obvious that the shapes of γ' particles in three oriented specimens are different when viewed along cross sections. In [001] specimens, cuboidal γ' particles are embedded in γ matrix phase. In [011] and [111] specimens, the γ' particles are rectangular and triangular, respectively. The average width of γ channel in [001] specimens is nearly 80 nm. In [011] specimens, the width of gable channel (shown in Fig. 1b) is approximately 85 nm. As for [111] samples, the γ channel width is calculated as $a\sin 54.7^\circ$, which is approximately 96 nm. The parameter a is the measured channel width on the (111) cross section as displayed in Fig. 1c. The 54.7° is the angle between (111) and (001) planes.

Fig. 1 Heat-treated microstructures of γ/γ' phases along the cross sections in **a** [001], **b** [011], **c** [111] specimens

3.2 Stress Rupture Properties

The stress rupture lives and elongations of [001], [011] and [111] specimens are displayed in Fig. 2. Each datum is an average of three specimens. It can be seen that the stress rupture anisotropy is shown at 1100 °C, but not so obvious compared with that at intermediate temperatures [8, 29]. The [001] specimens show the longest stress rupture life of 81.6 h, while the [111] specimens show the shortest life of 63.9 h. The [011] specimens display the intermediate life of 72.1 h. The [001] specimens present the elongation of 19.6%. The [011] samples show the shortest elongation of 14.7%. Though the [111] specimens show the matched elongation with [001] specimens, its standard deviation is much larger than that of [001] specimens, which means some specimens display longer elongations and the others exhibit shorter elongations.

Fig. 2 Stress rupture lives and elongations of [001], [011] and [111] specimens

3.3 Microstructures in Interrupted and Ruptured Specimens

Figure 3 shows the γ/γ' microstructures of [001], [011] and [111] specimens crept at 30 h and ruptured, respectively, at 1100 °C/150 MPa. In [001] specimens, rafted structures perpendicular to the loading direction were formed after 30 h creep (Fig. 3a). While for [011] and [111] specimens, γ' precipitates were merged to some extent after crept for

30 h, as displayed in Fig. 3c, e. The rapid rafting in [001] specimens is probably due to the higher stress level along horizontal channels, which is attributed to the superposition of coherency stress and applied tensile stress [9, 10]. Fully rafted microstructures are observed in the ruptured [011] and [111] specimens. In [011] specimens, the rafted structures are $\pm 45^\circ$ inclined to the loading direction when observed along (100) plane (Fig. 3d). In [111] specimens, the rafts are inhomogeneous with various directions when

Fig. 3 γ/γ' morphologies of three oriented samples tested at 1100 °C/150 MPa: **a** interrupted at 30 h and **b** after rupture of [001] specimens on (100) plane; **c** interrupted at 30 h and **d** after rupture of [011] specimens on (100) plane; **e** interrupted at 30 h and **f** after rupture of [111] specimens on (011) plane. Loading direction of all figures is along the vertical direction as marked in **a**

viewed along (01 $\bar{1}$) plane (Fig. 3f). Some rafts are inclined to the loading direction with small angles (region A), while others are nearly perpendicular to the loading direction (region B).

Figure 4 displays the fracture morphologies of different oriented specimens. In Fig. 4a, the [001] specimen exhibits a failure mode of microvoid coalescence. Rectangular and irregular facets can be seen on the fracture surface, on which creep or cast pores could be observed (Fig. 4b). Figure 4e and f shows that the microvoid accumulation fracture is prevailing in [011] specimens. In Fig. 4i, j, the [111] specimens show a mixed failure mode of microvoid coalescence and dendrite fracture. Unlike [001] specimens, most of the facets and micropores are irregular in [011] and [111] specimens.

In order to reveal the crack initiation and propagation, longitudinal sections of three oriented specimens were observed, as displayed in Fig. 4c, d, g, h, k. The dendrites are marked by the red lines. It can be seen that cracks mainly

initiate from pores in interdendritic regions, then propagate perpendicular to the loading direction and finally connect with each other along the interdendritic regions. Since the dendritic morphologies are different along the longitudinal sections of the three oriented specimens (Fig. 4c, g, k), cracks propagate along different interdendritic paths showing the distinct fracture morphologies (marked by arrows in Fig. 4c, g, k). For [001] specimens, the propagation paths are along the loading direction (Fig. 4c). For [011] and [111] specimens, the propagation paths are inclined to the loading directions (Fig. 4g, k). Moreover, numerous surface cracks are observed along the interdendritic regions of [111] oriented samples as shown in Fig. 4l.

3.4 Dislocation Configurations

3.4.1 Interfacial Dislocation Networks

Figure 5 shows the interfacial dislocation networks in [001], [011] and [111] specimens. It is obvious that the

Fig. 4 Fracture morphologies and longitudinal sections of **a–d** [001], **e–h** [011], **i–k** [111] specimens; **l** longitudinal surface of a [111] specimen

Fig. 5 Interfacial dislocation networks in **a, b** [001], **c, d** [011], **e, f** [111] specimens. Beam is [001] for [001] specimens; beam is [011] for [011] specimens; beam is close to $e[1\bar{1}2]$ and $f[111]$ for [111] specimens

networks in [001] specimens (Fig. 5a) are more regular than those in [011] and [111] specimens (Fig. 5c, e). In [001] specimens, dislocation networks are hexagonal (the inset in Fig. 5a) or rectangular (Fig. 5b), which could well impede dislocations from cutting into the γ' . In [011] and [111] specimens, the networks are irregular (Fig. 5c–f), so that dislocation pairs might easily cut into γ' as marked by black arrows.

3.4.2 Superdislocations in γ' Phase

Figure 6 shows the superdislocations observed in [001] specimens. It is found that two types of superdislocations are formed in [001] specimens. One is $a\langle 110 \rangle$ superdislocations,

which are short and arranged along $\langle 100 \rangle$ directions as shown in Fig. 6a. According to Ref. [18], these superdislocations are in a steeply inclined state so that only short segments can be observed. They lie on (100) or (010) plane and run parallel to [011] or [101] direction. They are formed by combinations of two $a/2\langle 110 \rangle$ dislocations with same Burgers vectors. Reaction can be given by [18]:

$$a/2\langle 101 \rangle + a/2\langle 101 \rangle = a\langle 101 \rangle \quad (1)$$

Another type of superdislocations is $a\langle 100 \rangle$ superdislocations as marked by triangles in Fig. 6b. They display the double-line contrast when viewed along [100] direction with g -vectors of 002, 020, 022 and $02\bar{2}$, as mentioned

Fig. 6 Superdislocations in [001] specimens: **a** $a\langle 110 \rangle$ superdislocations, $B=[001]$; **b** $a[100]$ superdislocations, $B=[100]$; **c** superdislocation node, $B=[001]$, **d** schematic illustration of the node

in Refs. [19–21]. The Burgers vector of the dislocations is identified as $a[100]$. These dislocations mainly lie on (100) plane with $[011]$ and $[01\bar{1}]$ directions. Different from $a\langle 110 \rangle$ superdislocations, such dislocations are formed by reactions of two matrix dislocations with different Burgers vectors as follows [21]:

$$a/2[101] + a/2[10\bar{1}] = a[100] \quad (2)$$

$$a/2[1\bar{1}0] + a/2[110] = a[100] \quad (3)$$

Besides the isolated superdislocations, superdislocation nodes are observed in [001] specimens (Fig. 6c). They are formed by reaction of two superdislocations cutting in γ' as illustrated in Fig. 6d. The reaction is shown as follows [2]:

$$a[011] - a[101] = a[\bar{1}10] \quad (4)$$

In [011] specimens, both $a\langle 110 \rangle$ and $a\langle 100 \rangle$ superdislocations are observed (Fig. 7). Figure 7a shows the $a\langle 110 \rangle$ superdislocations. After cutting into γ' , these dislocations usually split into two superpartials with the same Burgers vectors, which are coupled with antiphase boundaries (APBs). Figure 7b–d displays the morphologies of $a\langle 100 \rangle$ superdislocations. They show the double-line contrast when viewed along [100] direction. The Burgers vector of these dislocations is identified as $a[100]$, based on the diffraction contrast analyses [19–21]. They are formed according to Eq. (2, 3). Similar with $a[100]$

superdislocations in [001] specimens, these dislocations mainly lie on (100) plane and along $\langle 011 \rangle$ directions.

In [111] specimens, superdislocation arrays are observed along the cross (Fig. 8a) and longitudinal (Fig. 8b) sections, which indicates that one slip system may dominate the deformation. In Fig. 8c, d, masses of dislocations are piled up around the matrix and interfacial hexagonal networks are formed. In the vicinity of the interface, dislocation pairs cut into γ' phase and form the superdislocation networks as marked by white arrows in Fig. 8c, d. These superdislocation networks can be formed following Eq. (4). The networks in γ and γ' phase indicate that this region is subjected to high stress and strain concentration.

To investigate the Burgers vectors of superdislocations in [111]-oriented specimens, eleven typical superdislocations have been observed under six two-beam conditions and are displayed in Fig. 9. Several long parallel dislocations are observed, which are labeled as 4, 5, 9 and 11. And there also exist short segments as marked by 1–3, 6–8 and 10. A summary of the dislocation visibilities for various diffraction conditions is provided in Table 1. It is found that most of the segments are $a[011]$ dislocations as marked by black arrows, and the dislocations 3 and 6 have a Burgers vector of $a[10\bar{1}]$. With respect to the dislocation 7, it is in contrast in all two-beam conditions except in $g = 2\bar{2}0$; thus, the Burgers vector of dislocation 7 is deduced to be $a[001]$. Overall, most of the superdislocations are identified as $a\langle 110 \rangle$ type and only a few $a[001]$ superdislocations are observed.

Fig. 7 Superdislocations in [011] specimens: **a** $a\langle 110 \rangle$ superdislocations in γ' phase, **b** $a[100]$ superdislocations viewed along [100] direction, **c**, **d** magnified figures of $a[100]$ superdislocations in **b**. All images were taken close to the [100] zone axis

Fig. 8 Superdislocations in [111] specimens: **a**, **b** superdislocation arrays viewed along the cross ($B=[111]$) and longitudinal sections ($B=[\bar{1}\bar{1}2]$), respectively, **c**, **d** superdislocation networks in γ' phase, $B=[\bar{1}\bar{1}2]$

Fig. 9 TEM micrographs of superdislocation 1–11 under six two-beam conditions in $[111]$ -oriented specimens: **a** $g = \bar{1}11$, **b** $g = 02\bar{2}$, $B = [211]$; **c** $g = \bar{2}02$, **d** $g = 1\bar{1}1$, $B = [121]$; **e** $g = 2\bar{2}0$, **f** $g = 11\bar{1}$, $B = [112]$. Foil normal is close to $[111]$ direction

Table 1 Summary of dislocation contrast analysis for superdislocation 1–11 in $[111]$ specimens in Fig. 9

Dislocation	1	2	3	4	5	6	7	8	9	10	11
$g = \bar{1}11$	V	V	V	V	V	V	V	V	V	V	V
$g = 02\bar{2}$	I	I	V	I	I	V	V	I	I	I	I
$g = \bar{2}02$	V	V	V	V	V	V	V	V	V	V	V
$g = 1\bar{1}1$	I	R	I	I	I	I	V	I	I	I	I
$g = 2\bar{2}0$	V	V	V	V	V	V	I	V	V	V	V
$g = 11\bar{1}$	I	I	V	I	I	V	V	I	I	I	I
Deduced b	$a[011]$	$a[011]$	$a[10\bar{1}]$	$a[011]$	$a[011]$	$a[10\bar{1}]$	$a[001]$	$a[011]$	$a[011]$	$a[011]$	$a[011]$

V visible, I invisible, R residual contrast, $g \cdot b = 0$ but $g \cdot (b \times u) \neq 0$ conditions

4 Discussion

4.1 Effects of Crack Propagation on Stress Rupture Anisotropy

During high-temperature creep, ultimate failure is closely related to the cast pores, creep pores and TCP-related pores in the form of cavitation [13, 28], especially for the specimens machined from large-scale SX slabs [30, 31]. The relatively different distribution of pores in samples of different orientations results in the fluctuations of rupture life and elongation [13], which are consistent with the results in Fig. 2. In the present work, no TCP was observed; thus, the cast and creep pores are proposed to be the crack sources. Microcracks mainly initiate in the interdendritic regions

from the micropores [30, 32]. They connect with each other along the interdendritic regions and come to the final rupture [30]. For specimens with different orientations, distinct morphologies of dendrite are shown in Fig. 4c, g, k, which match well with the corresponding fracture morphologies (marked by arrows in Fig. 4c, g, k). Besides, the alignment of cast pores is related to the dendritic morphology, which can also affect the propagation of the cracks and final fracture [27].

Based on the fracture morphologies and longitudinal sections of the ruptured specimens (Fig. 4), the alignment of micropores and crack propagation paths in $[001]$, $[011]$ and $[111]$ specimens is illustrated in Fig. 10. In $[001]$ specimens, the pores tend to align along the loading direction. It means that cracks have to propagate by cutting through the primary dendrites (marked by dashed line in Fig. 10a) or along the loading directions which can be observed in Fig. 4c marked

Fig. 10 Schematic illustration of the alignment of micropores and crack propagation paths in **a** [001], **b** [011], **c** [111] specimens

by arrow. Mälzer et al. [27] thought that for specimens with preexisting pores parallel to the loading direction, it was difficult for microcracks to link up, thus leading to the late onset of tertiary creep. In [011] samples, the dendrites and alignment of pores are around $\pm 45^\circ$ to the applied stress (Fig. 10b). This configuration can help the cracks to extend along [010] and [001] directions in interdendritic region as marked by arrows in Figs. 4g and 10b. While in [111] specimens, pores distribute in three $\langle 001 \rangle$ directions in interdendritic regions, so that the microcracks can connect along [001], [010] and [100] directions (Fig. 10c), which may result in the dendrite fracture as displayed in Fig. 4i. Moreover, due to the complicated arrangement of dendrites in [111] specimens, cracks tend to propagate along the interdendritic regions on the sample surfaces as shown in Fig. 4i. More crack propagation paths and surface cracks may be responsible for the faster creep failure of [111] specimens compared with [001] and [011] specimens. Besides the pores in interdendritic regions, Epishin et al. [17] found that new pores might nucleate at the γ/γ' interfaces during high-temperature creep, but these pores were smaller and homogeneously distributed in the specimens, thus might not affect the stress rupture anisotropy.

4.2 Effects of Microstructure and Dislocations on Stress Rupture Anisotropy

Directional coarsening or rafting takes place prevalently at high temperatures in nickel-based SX superalloys. It is driven by different chemical potentials in different matrix channels [24]. The morphology of rafted structures depends on the external stress, lattice misfit and crystal orientation [22, 33, 34]. It is believed that rafted structures strongly affect the dislocation movement in matrix channel [23, 26, 34]. The effects of different rafted structures on dislocation

movement are summarized and illustrated in Fig. 11. For [001] specimens, typical *N*-type rafted structures are formed and the vertical γ channels are eliminated. The slip and climb of matrix dislocations are hindered by rafted γ' phase, and dislocations can only move in horizontal channels, which lead to the higher creep resistance [12, 13, 22]. For [011] specimens, the rafted structures are $\pm 45^\circ$ inclined to the loading direction. Dislocations can easily glide and climb along the $\pm 45^\circ$ rafted structures, and thus lower the creep resistance. As for [111] oriented samples, three matrix channels are theoretically equivalent in chemical potential, which could result in the symmetrical deformation in each matrix channel [35], but a small-angle deviation from precious [111] direction is inevitable [11]. The equivalence of chemical potential in three matrix channels will break down and lead to the rafts with preferred directions. As shown in Figs. 3f and 11c, rafts in region A are inclined to the loading direction with a small angle, while rafts in region B are perpendicular to the loading direction. Matrix dislocations can glide and climb in region A and will be hindered by rafted γ' in region B. Dislocation pile-up takes place at the joints of regions A and B and results in the stress concentration, which can be clearly observed in Fig. 12. Local stress concentration caused by irregular rafts suggests that heterogeneous deformation takes place thus deteriorating the property of [111] specimens [22]. Moreover, the inhomogeneous rafted structures vary from specimen to specimen, which may have different effects on dislocation movement and increase the instability of elongation in [111] specimens.

During creep, the interfacial dislocation networks are formed concurrently with rafts. There is evidence that the stability and integrity of the dislocation networks have great effect on creep properties [15, 33]. The networks in [001] specimens (Fig. 5a) are more regular than those in [011] and [111] specimens (Fig. 5c, e). This can be rationalized

Fig. 11 Effects of different rafts on dislocation movement: **a** slip and climb of dislocations are hindered by *N*-type rafts in [001] specimens; **b** dislocations can easily glide and climb along the $\pm 45^\circ$ rafts in [011] specimens; and **c** dislocations are piled up at the joints of regions A and B in [111] specimens

Fig. 12 Dislocations are piled up at the joints of regions A and B, in [111] ruptured specimens, $B = [\bar{1}\bar{1}2]$

by the different activated slip systems for SX alloys when crept along different orientations [22]. Theoretically, eight $\langle 110 \rangle$ {111} slip systems with Schmid factor of 0.41 can be equally activated in [001] specimens. For [001] specimens with misorientations, most of these slip systems are activated in the early creep stage [22, 24]. Matrix dislocations glide and cross-slip in matrix channels and react to form dense dislocation networks at γ/γ' interface [9] as shown in Fig. 5a, b. For [011] specimens, the number of activated slip systems is half of that in [001] specimens. Only (111)[1 $\bar{1}$ 0], ($\bar{1}\bar{1}$ 1)[110], (111)[10 $\bar{1}$] and ($\bar{1}\bar{1}$ 1)[101] slip systems with Schmid factor of 0.41 can be activated in (010) or (001) channels [34]. Besides, the cross-slip in [011] specimens is difficult due to the relative alignment of slip systems [9]. Therefore, less dislocation interactions occur and the dislocation networks may be coarser than those in [001] specimens. For [111] specimens, six $\langle 110 \rangle$ {111} slip systems with Schmid factor of 0.27 may be activated, but the value

of the Schmid factor is lower than those in [001] and [011] specimens, which means higher stress is required to activate the slip systems [10]. Besides, small misorientation can further reduce the number of the activated slip systems. Thus in [111] specimens, networks are not as regular as those in [001] specimens. Overall, the regular dislocation networks in [001] specimens may well hinder the matrix dislocations from cutting into γ' and strengthen the alloy, but the irregular networks in [011] and [111] specimens may be related to the reduction of creep resistance.

During the later steady and tertiary creep stages, superdislocation cutting rafted γ' is the dominant mechanism that controls the recovery [19, 36] and softening processes [37]. The creep strain rate can be evaluated as follows [38]:

$$\dot{\epsilon}_c = \rho_m b v \quad (5)$$

where ρ_m is the mobile dislocation density, b is the magnitude of Burgers vectors, and v is the average velocity of the mobile dislocations. That is to say, the density, type and mobility of superdislocations in γ' phase control the creep rates [25, 37]. In this paper, the types of superdislocation in [001], [011] and [111] specimens are analyzed in Sect. 3.4.2. Two types of superdislocations, $a\langle 110 \rangle$ and $a[100]$ are prevalently formed in [001] and [011] specimens, while lots of $a\langle 110 \rangle$ and a few $a[001]$ superdislocations are observed in [111] specimens. The well known $a\langle 110 \rangle$ superdislocations can simply glide on {111} planes [17, 39], but the $a[100]$ superdislocations can only move by pure climb or the coupled glide and climb of two $a/2\langle 110 \rangle$ superpartials in [001] and [011] specimens due to the lack of resolved stress for slip [17, 24, 36]. Besides, the magnitude of $a[100]$ superdislocations ($|b| = a$) is smaller than that of $a\langle 110 \rangle$ superdislocations ($|b| = \sqrt{2}a$). Considering Eq. (5), the $a[100]$ superdislocations with lower mobility and magnitude are favorable for SX alloys to maintain a low creep strain rate

[25, 40]. Therefore, the formation of $a[100]$ superdislocations in $[001]$ and $[011]$ specimens may be responsible for a low creep strain rate. Even though a few $a[001]$ superdislocations are observed in γ' phase in $[111]$ specimens, these dislocations can glide on (110) plane with a Schmid factor of 0.47 or glide on (010) plane with a Schmid factor of 0.33 [41]. The $a\langle 110 \rangle$ and $a[001]$ dislocations possess high mobility in $[111]$ specimens, which indicates that a higher strain rate can be obtained.

Considering the deformation mechanism, both interfacial dislocation networks and superdislocations are observed in three oriented specimens, but the morphology of dislocation networks and type of superdislocations are distinct. The regular dislocation networks may enhance the creep resistance of $[001]$ specimens [26], and the $a[100]$ superdislocations with low mobility may be responsible for the low creep strain rate of $[001]$ and $[011]$ specimens [25, 40]. Whereas, further work is required to quantitatively study the correlation between creep performance and interfacial dislocation spacings of different oriented specimens based on a unified scale.

5 Conclusions

In the present work, the stress rupture anisotropy of a 3rd-generation nickel-based SX superalloy DD33 was investigated at 1100 °C/150 MPa. The following conclusions can be drawn.

1. Under 1100 °C/150 MPa, the stress rupture anisotropy is observed. The rupture life decreases in the sequence of $[001]$ specimens, $[011]$ specimens and $[111]$ specimens.
2. In $[001]$ specimens, the slip and climb of dislocations are well hindered by N -type rafts, and γ' cutting is prevented by the regular dislocation networks. Moreover, less crack propagation paths are observed. All these lead to the longest rupture life of $[001]$ specimens.
3. For $[011]$ specimens, alignment of $\pm 45^\circ$ rafts covered with irregular networks may result in the worse stress rupture property than that of $[001]$, while the similar cutting mechanisms (both $a\langle 110 \rangle$ and $a[100]$ superdislocations) may reduce the creep anisotropy of $[001]$ and $[011]$.
4. The shortest stress rupture life of $[111]$ specimens may be interpreted mainly by the maximum number of crack propagation paths, as well as the irregular rafted structures and superdislocations with high mobility.

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References

- [1] R.C. Reed, *The Superalloys: Fundamentals and Applications* (Cambridge University Press, Cambridge, 2006)
- [2] G.S. Hillier, *Doctoral Thesis*, (Cambridge University, Cambridge, 1984)
- [3] T.M. Pollock, A.S. Argon, *Acta Metall. Mater.* **40**, 1 (1992)
- [4] H.B. Long, S.C. Mao, Y.N. Liu, Z. Zhang, X.D. Han, *J. Alloys Compd.* **743**, 203 (2018)
- [5] J.C. Han, *Int. J. Rotating Mach.* **10**, 443 (2004)
- [6] R.A. Mackay, R.D. Maier, *Metall. Trans. A* **13**, 1747 (1982)
- [7] M. Segersäll, J.J. Moverare, K. Simonsson, S. Johansson, in *Superalloys*, ed. by E.S. Huron, R.C. Reed, M.C. Hardy, M.J. Mills, R.E. Montero, P.D. Portella, J. Telesman (TMS, Warrendale, 2012), p. 215
- [8] Q.Z. Chen, D.M. Knowles, *Mater. Sci. Eng., A* **356**, 352 (2003)
- [9] V. Sass, U. Glatzel, M. Feller-Kniepmeier, *Acta Mater.* **44**, 1967 (1996)
- [10] V. Sass, M. Feller-Kniepmeier, *Mater. Sci. Eng., A* **245**, 19 (1998)
- [11] J.L. Liu, T. Jin, X.F. Sun, J.H. Zhang, H.R. Guan, Z.Q. Hu, *Mater. Sci. Eng., A* **479**, 277 (2008)
- [12] T.P. Gabb, S.L. Draper, D.R. Hull, R.A. Mackay, M.V. Nathal, *Mater. Sci. Eng., A* **118**, 59 (1989)
- [13] R.C. Reed, N. Matan, D.C. Cox, M.A. Rist, C.M.F. Rae, *Acta Mater.* **47**, 3367 (1999)
- [14] A. Epishin, T. Link, U. Brückner, P.D. Portella, *Acta Mater.* **49**, 4017 (2001)
- [15] J.X. Zhang, T. Murakumo, H. Harada, Y. Koizumi, *Scr. Mater.* **48**, 287 (2003)
- [16] J.X. Zhang, J.C. Wang, H. Harada, Y. Koizumi, *Acta Mater.* **53**, 4623 (2005)
- [17] A. Epishin, T. Link, *Philos. Mag.* **84**, 1979 (2004)
- [18] J.X. Zhang, T. Murakumo, Y. Koizumi, T. Kobayashi, H. Harada, *Acta Mater.* **51**, 5073 (2003)
- [19] P.M. Sarosi, R. Srinivasan, G.F. Eggeler, M.V. Nathal, M.J. Mills, *Acta Mater.* **55**, 2509 (2007)
- [20] T. Link, A. Epishin, M. Klaus, U. Brückner, A. Reznicek, *Mater. Sci. Eng., A* **405**, 254 (2005)
- [21] M. Huang, L.C. Zhuo, J.C. Xiong, J.R. Li, J. Zhu, *Philos. Mag. Lett.* **95**, 496 (2015)
- [22] Y. Su, S.G. Tian, H.C. Yu, D.L. Shu, S. Liang, *Mater. Sci. Eng., A* **668**, 243 (2016)
- [23] H.C. Yu, Y. Su, N. Tian, S.G. Tian, Y. Li, X.F. Yu, L.L. Yu, *Mater. Sci. Eng., A* **565**, 292 (2013)
- [24] L.A. Jácome, P. Nörtershäuser, J.K. Heyer, A. Lahni, J. Frenzel, A. Dlouhy, C. Somsen, G. Eggeler, *Acta Mater.* **61**, 2926 (2013)
- [25] L.A. Jácome, P. Nörtershäuser, C. Somsen, A. Dlouhy, G. Eggeler, *Acta Mater.* **69**, 246 (2014)
- [26] G.L. Wang, J.L. Liu, J.D. Liu, T. Jin, X.F. Sun, X.D. Sun, Z.Q. Hu, *J. Mater. Sci. Technol.* **32**, 1003 (2016)
- [27] G. Mälzer, R.W. Hayes, T. Mack, G. Eggeler, *Metall. Mater. Trans. A* **38**, 314 (2007)
- [28] R.C. Reed, D.C. Cox, C.M.F. Rae, *Mater. Sci. Eng., A* **448**, 88 (2007)
- [29] Y.F. Li, L. Wang, G. Zhang, J. Zhang, L.H. Lou, *Mater. Sci. Eng., A* **760**, 26 (2019)
- [30] M. Huang, G. Zhang, D. Wang, J.S. Dong, L. Wang, L.H. Lou, *Acta Metall. Sin. (Engl. Lett.)* **31**, 887 (2018)
- [31] X.W. Jiang, H. Li, *Foundry Technol.* **33**, 807 (2012)

- [32] L. Liu, J. Meng, J.L. Liu, H.F. Zhang, X.D. Sun, Y.Z. Zhou, *Acta Metall. Sin. (Engl. Lett.)* **32**, 381 (2019)
- [33] T.M. Pollock, R.D. Field, in *Dislocations in Solids*, ed. by F.R.N. Nabarro, M.S. Duesbery (Elsevier, Amsterdam, 2002)
- [34] Y. Su, S.G. Tian, H.C. Yu, L.L. Yu, *Scr. Mater.* **93**, 24 (2014)
- [35] E.R. Golubovskii, I.L. Svetlov, *Strength Mater.* **34**, 109 (2002)
- [36] R. Srinivasan, G.F. Eggeler, M.J. Mills, *Acta Mater.* **48**, 4867 (2000)
- [37] Y.L. Tang, M. Huang, J.C. Xiong, J.R. Li, J. Zhu, *Acta Mater.* **126**, 336 (2017)
- [38] F.R.N. Nabarro, H.L. De Villiers, *Physics of Creep and Creep-Resistant Alloys* (Taylor and Francis Ltd., London, 1995)
- [39] Q.Z. Yue, L. Liu, W.C. Yang, C. He, D.J. Sun, T.W. Huang, J. Zhang, H.Z. Fu, *J. Mater. Sci. Technol.* **35**, 752 (2019)
- [40] X.G. Wang, J.L. Liu, T. Jin, X.F. Sun, Y.Z. Zhou, Z.Q. Hu, J.H. Do, B.G. Choi, I.S. Kim, C.Y. Jo, *Mater. Sci. Eng., A* **626**, 406 (2015)
- [41] T. Link, C. Knobloch, U. Glatzel, *Scr. Mater.* **40**, 85 (1999)