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Deformation localization in a single crystal nickel-based superalloy oriented for multiple slip

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ABSTRACT

The deformation and fracture in tension of a single crystal nickel-based superalloy were studied with relation to the crystallographic orientations. The [011] oriented sample exhibited significant heterogeneous deformation by asymmetric necking and deformation bands. On the contrary, the [111] one deformed in a macroscopically homogeneous way. Microstructure observation and EBSD analysis showed that those deformation bands in the [011] oriented sample were formed by concentration of particle shearing and led to lattice rotation of about 15°. Slip systems operating in the matrix channels caused the huge difference in the deformation behavior of the two orientations. Some related studies by other researchers carried out at different temperatures or stress states were also discussed.

1. Introduction

Deformation localization was widely studied due to its close connection with fracture. Localized deformation was common in ductile metals and alloys, especially in single crystals [1,2]. Deformation localization was usually manifested as the onset of diffuse necking and localized shearing, which gave way to a concentration of deformation in bands and led directly to the final fracture. Precipitate strengthened alloys were more likely to form intense macroscopic shear bands [3,4]. The deformation traces and stress response had been systematically studied in single crystals containing coherent particles, such as aluminum-zinc, aluminum-copper and copper-beryllium alloys [5,6]. The occurrence of localized shearing was generally associated with a critically low value of strain hardening rate on the active slip plane to the current stress level and the misorientation in the shear bands which led to geometrical softening [7–9].

Single crystal nickel-based superalloys were precipitate strengthened by the coherent γ′ phase based on Ni_3Al. The complicated deformation behaviors of nickel-based superalloys were largely attributed to the $L1_2$ structure of the second phase [10–14]. Deformation localization by lattice rotation and macroscopic shear bands had been observed in single crystal nickel-based superalloys even when they were oriented for symmetric multiple slip [15–18]. Although the [001] orientation were the most commonly used due to its comprehensive mechanical property, superalloys might also be subjected to external stress along other directions. Early studies found that deformation of the [011] oriented nickel based superalloys was very special. It often showed the largest tension/compression asymmetry in yield stress, a nearly unchanged yield stress from room to intermediate temperature, lattice rotation, and intense shear bands. In comparison, the [111] orientation often deformed in a homogeneous way and its yield stress decreased rapidly with increasing temperature [19,20]. Moreover, because of the low mobility of the superdislocation in the γ′ phase, fracture features of the superalloys were also found to be different from normal FCC metals with cleavage along {111} and {001} planes from room to elevated temperatures [21,22]. Conventional FCC metals could also develop cracks on the {111} plane, but usually not in a cleavage way [22–24].

This work mainly focused on the deformation localization in a single crystal nickel-based superalloy in [011, 111] directions, both of which were favorable to symmetric multiple slip for typical FCC metals. This paper tried to give detailed description and analysis of the fracture characteristics, orientation distribution and dislocation configuration with respect to the crystallographic orientations. Some related studies by other researchers carried out at different temperatures or stress states were also discussed.

2. Experimental

The nickel based single crystal superalloy DD6 was investigated. It was a second generation single crystal superalloy developed by China. Solution and aging treatment was carried out on the alloy under an optimum condition as follows: 1290 °C/1 h + 1300 °C/2 h + 1315 °C/

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Table 1
Chemical composition of the single crystal nickel-based superalloy chosen for this work (wt%).

C	Cr	Ni	Co	W	Mo	Al
0.001–0.04	3.8–4.8	Bal	8.5–9.5	7.0–9.0	1.5–2.5	5.2–6.2
Ti	Fe	Nb	Ta	Re	Hf	B
≤0.10	≤0.30	0–1.2	6.0–8.5	1.6–2.4	0.05–0.15	≤0.02

Table 2
Tensile properties of the [011, 111] orientations at room temperature.

	σ_y (MPa)	σ_b (MPa)	δ (%)
111	1035	1527	16
011	966	1150	17

4 h + air cooling + 1120 °C/4 h + air cooling + 870 °C/32 h + air cooling. Chemical composition of this alloy was shown in Table 1.

Cylinder specimens oriented in [011, 111] directions were tensioned to fracture under a strain rate of $10^{-3}\ s^{-1}$ at room temperature. Microstructure observation and EBSD analysis were carried out under an Apollo-300 field-emission scanning electron microscope. For EBSD analysis, thin sections were cut parallel to the loading axis from the fractured specimens. They were firstly grinded by abrasive papers and then electron polished in a 90% C_3H_6O + 10% $HClO_4$ solution for 1 min under a voltage of 20 V. EBSD analysis was carried out to investigate the orientation distribution. Euler angles ($\phi_1 \phi \phi_2$) of each tested points were obtained. By multiplying the following conversion matrixes, the accurate crystallographic orientation vectors could be calculated. The three row vectors of the product were corresponding to the loading axis, the transverse direction and the surface normal respectively.

$$\begin{bmatrix} \cos\varphi_1 & -\sin\varphi_1 & 0 \\ \sin\varphi_1 & \cos\varphi_1 & 0 \\ 0 & 0 & 1 \end{bmatrix} \begin{bmatrix} 1 & 0 & 0 \\ 0 & \cos\varphi & -\sin\varphi \\ 0 & \sin\varphi & \cos\varphi \end{bmatrix} \begin{bmatrix} \cos\varphi_2 & -\sin\varphi_2 & 0 \\ \sin\varphi_2 & \cos\varphi_2 & 0 \\ 0 & 0 & 1 \end{bmatrix}$$

Transmission electron microscopy (TEM) was performed by a F20 field-emission transmission electron microscope under a voltage of 200 kV. The TEM samples were prepared by thin slices extracted from the failure specimens with different distances to the fracture surface. They were firstly mechanical grinded to a thickness of around 30 μm and punched into round foils with a diameter of 3 mm. Final thinning to the thickness of electron transparency was carried out using a Res101 Ion Milling System.

3. Results and discussion

3.1. Tensile properties and fractograph

Tensile stress-strain curves were shown in Fig. 1. The [111] oriented sample exhibited higher yield stress (σ_y) and fracture stress (σ_b) than the [011] one. These results were concluded in Table 2. In addition, the [011] orientation showed a rather week tendency of work hardening in comparison with the other one, indicating that there should be some softening mechanism during its tensile process. Serrated flow of the [011] oriented superalloy caused by dynamic strain aging in another study was not observed here. It was probably due to the much slower strain rate of 10^{-5}/s and higher temperature of 500 °C adopted in their research [19]. An investigation of another second generation single crystal superalloy showed that, from room temperature to around 700 °C, the yield stress kept almost unchanged for the [011] orientation and decreased for the [111] one. This tendency was attributed to the motion of superdislocations in the γ′ phase which was controlled by cross slip of the dislocation core from octahedral slip planes to the cube planes. The theory of superdislocation in Ni_3Al had been systematically studies in early researches.

Fig. 2 made a sharp contrast in the fracture morphology of the two orientations. The fractured [011] sample exhibited severe heterogeneous deformation by necking and macroscopic defamation bands. These bands were widely spaced and did not run through the whole section. Their propagation was interrupted by the activation of another slip system. However, the traces of this slip system were too fine to be seen in Fig. 2a and were shown by optical microscopic image in Fig. 2b. For the [111] orientation, the deformation was macroscopically homogeneous as shown in Fig. 2c. Some other studies carried out at intermediate temperature or under a compression stress state showed similar deformation features. The rise in temperature or the change of the stress state mainly influenced the yield stress. In Mikael Segersäll and Johan J. Moverare's work, macroscopic deformation bands of the [011] oriented crystals were assumed to be bundles of fine slip lines cutting the second phase, however, continuous shearing of the γ′ phase was not observed in those bands in their interrupted compression experiment and was attributed to the interruption of deformation at small plastic strain [19]. In our study, the microstructure obtained at the boundary of the deformation bands verified their assumption. It was seen in Fig. 3 that there were closely packed slip lines cutting the second phase and they were basically parallel to the macroscopic deformation bands. No trace of particle shearing was found in the other area

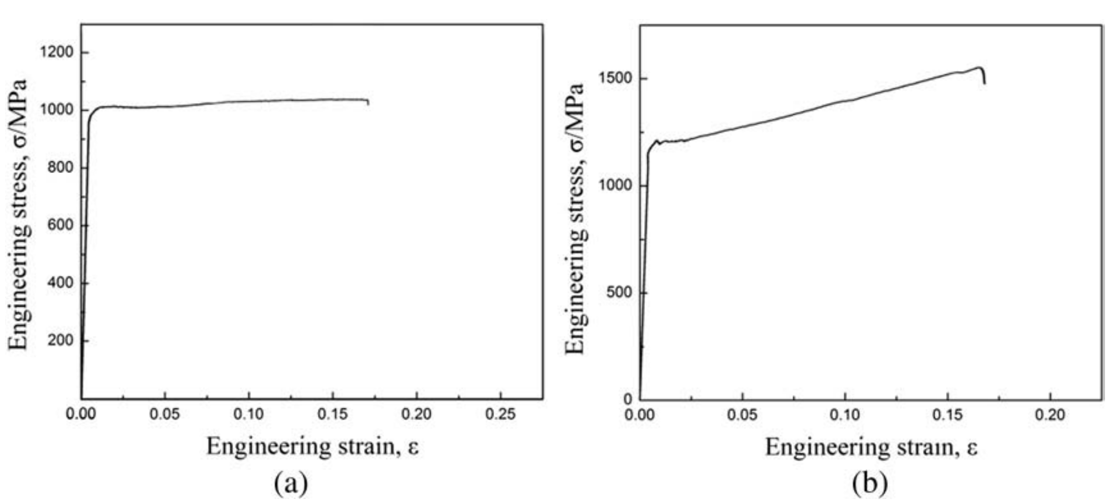

Fig. 1. Tensile strain-stress curves. (a) [011] oriented sample. (b) [111] oriented sample. It appeared that the [011] one had a rather weak tendency of strain hardening, indicating that there should be some softening mechanism during its tensile process.