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Different roles of stacking fault energy and diffusivity in the creep performance of nickel-based single-crystal superalloys

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Abstract

Two nickel-based single-crystal (SC) superalloys (designated T6 and T13) were investigated in order to reveal the effects of stacking fault (SF) energy and diffusion on their high-temperature creep behavior. In this study, the effect of dislocation spacing on creep was the same between T6 and T13. The microstructure and deformation rate attained prior to 1% creep were investigated in detail. Several differences were detected in the two superalloys, the reasons for these differences were discussed. The results suggested that the SF energy of the γ matrix was the key factor affecting development of the dislocation network. Because the matrix of the T6 superalloy has lower SF energy than T13, the cross-slipping of dislocation was more difficult in the early stages of creep, which resulted in lower levels of dislocation propagation and of formation of the dislocation network. When creep had entered the steady stage, the key process controlling high-temperature creep was atomic migration in the γ matrix, and the creep rupture life was lengthened by reducing the penetration of dislocations into the γ′ raft and slowing down its topological inversion.

1. Introduction

The creep performance of nickel-based single-crystal (SC) superalloys for turbine blades has been continually improved through optimization of the composition of the superalloys and the use of directional solidification techniques [1, 2]. The precipitation of the γ′ phase and the solution of refractory elements are very important for strengthening alloys [3–7]. In addition, the interaction between dislocations can also significantly increase the creep performance of superalloys [3–7]. Neatly arranged γ′ particles restrict dislocations moving into the narrow orthogonal matrix channels [1, 8, 9]. The movement of dislocations is further reduced by the presence of refractory elements in the matrix channel, especially W, Mo, Re, and Ru [10, 11].

Nickel-based SC superalloys are typical alloys strengthened with a high fraction of γ′ precipitates [1, 2]. Al and Ta are the two important elements for γ′ precipitation. Mo and W are commonly added to superalloys to elevate the resistance of the γ and γ′ phases synchronously due to their comparable partition behavior [10, 12]. In order to improve their creep resistance at high temperatures, two important elements, Re and Ru, are usually added [4, 13, 14], a practice that has been adopted as the criterion for categorizing the new generation of SC superalloys [1]. Re has the lowest diffusivity in Ni of all metals and can significantly improve the creep rupture life at high temperatures [15, 16]. However, Re strongly promotes the formation of the topologically close-packed (TCP) phase [17], which is harmful to creep rupture life [18, 19]. The addition of Ru can inhibit the precipitation of the TCP phase during superalloy service [13, 18]. Additionally, the low diffusivity of Ru [20], which is approximately equivalent to that of W, can also improve the creep resistance effectively [21]. Detailed work on this has been carried out by the Japan National Institute for Material Science (NIMS), who turned their attention to crystal faults that cause microstructure degradation and developed several new SC superalloys, such as TMS-138 [6, 7, 22], TMS-196 [23, 24], and TMS-238 [25]. It was discovered that the addition of Mo, Re, and Ru caused more negative misfitting that favored densification of the dislocation network, which renders
dislocation inactive and thereby increases creep resistance at high temperatures [6, 7, 22]. The minimum creep rate showed an approximately linear relationship with dislocation spacing [6].

It is undisputed that the addition of alloying elements can also change the stacking fault energy synchronously [15, 26, 27], except for the effect on lattice misfit and diffusivity. SF energy and diffusivity are two important factors for enhancing creep performance [13, 28, 29]. NIMS focused on the relationship between dislocation spacing and minimum creep rate, while less attention was paid to the effect of decreasing SF energy. There are two major reasons for this: firstly, the SF shear of the γ′ phase has been documented mainly following intermediate- and low-temperature creep. Since SF has rarely been observed in high-temperature creep, not enough attention has been paid to the effects of SF energy on high-temperature creep. Secondly, the SF energy of the γ matrix is difficult to measure experimentally due to the γ channel being very narrow and the dislocation density high. Despite these two considerations, it is suggested that the SF energy has an important effect on high-temperature creep. There are many studies of the effects of SF energy on intermediate- and low-temperature creep, but the effects of SF energy on high-temperature creep are unknown. High-temperature creep is the main damage mode of single crystal blade, so it is necessary to study the role of SF energy in it. The characteristics of the high-temperature creep curve are numerous before creep attains 1% plastic deformation [30, 31]. This is the result of the synergistic effect of lattice misfit, diffusivity, and SF energy. Although the roles of diffusion in high-temperature creep have been recognized, the diffusion have a different effect on creep performance relative to SF energy. By comparing creep performance and development of the dislocation network for high-temperature creep, the different roles of SF energy and diffusivity can be illustrated in detail.

### 2. Materials and methods

Two nickel-based SC superalloys were used, the nominal compositions of which are listed in Table 1. In order to meet the requirements of the investigation, the experimental superalloys were designed using JMatPro® 7.0 software (https://sentesoftware.co.uk). The main physical properties in thermodynamic equilibrium were optimized by adjusting the input composition. The lattice misfits of γ and γ′ at 1100 °C are approximately equal, although T13 has a greater Mo content than T6. In addition, T6 has a lower SF energy due to the greater contents of Re and Ru. The microstructure stability of the two superalloys was satisfactory, and the calculated TCP phase fraction was less than 5%.

|       | Co | Cr | Al+Ta | Hf | Mo | W + Re | Ni |
|-------|----|----|-------|----|----|--------|----|
| T6    | 12 | 3  | 12.6  | 0.1| 0.8| 17.2    | Bal.|
| T13   | 12 | 2.5| 13.7  | 0.1| 3  | 13.5    | Bal.|

The master alloys were melted in a vacuum induction furnace and cast into ingots. SC bars were grown in a Bridgman directional solidification furnace at a constant withdrawal rate of 6 mm min⁻¹. A homogenization treatment was carried out as follows: 1310 °C, 2 h + 1315 °C, 3 h + 1320 °C, 5 h + 1325 °C, 8 h + 1330 °C, 20 h followed by air cooling (AC). First-step aging selected 1150 °C, 4 h (AC) and second-step aging was performed at 870 °C, 24 h (AC). The rotary x-ray diffraction method was used in this study to detect the difference in orientation, and its physical basics and mechanism have been described in detail in Reference [32]. The sample with an axial orientation deviating in the [001] crystal direction above 5° was excluded from the creep sample in order to reduce the impact of the difference in crystal orientation. The samples were tensile crept at 1100 °C and 150 MPa. Except for rupture, the creep experiment was also interrupted at 1% plastic deformation and 20 h to document the microstructure of the primary and steady creep. In addition, a 10 h creep interruption was added for T13. The external loading was not removed during cooling until the temperature dropped below 300 °C.

The samples for microstructure detection were cut from the middle of the rods and mechanically polished using the standard metallographic method. The samples were chemically etched in a solution of 20 g of CuSO₄ + 100 ml of HCl + 80 ml of H₂O + 5 ml of H₂SO₄. The characterization of the dislocation microstructure by scanning electron microscopy (SEM) is unique to nickel-based superalloys, the physical basics and application of which have been discussed in Reference [33]. In the present study, the SEM method was also used to characterize the dislocation line near the γ′ surface. The dislocation spacing was measured in the SEM microstructure of the 1% plastic deformation, and the lattice misfit was calculated based on these dislocation spacings. The diffusivity of the γ matrix was estimated using the method described in Reference [8] with the data for the Ni-X (X: Re, Ta, W, Co, Cr, Mo, Ru, or Al) binary alloy system [16, 34–36]. The formation of the SF
approach to lattice transforms from face-centered cubic (FCC) to hexagonal close-packed (HCP). The energy barrier of the transformation can represent the degree of difficulty of SF formation. The Gibbs free energy of lattice transformation was calculated using the Thermal-Calc software (https://thermocalc.com/products/thermo-calc).

3. Results

3.1. Creep behavior

Figures 1(a)–(f) shows the creep behavior of the superalloys tested at 1100 °C/150 MPa. The creep strain versus time curves (figures 1(a), (c), and (e)) belong to creep rupture, 1% plastic deformation, and 20 h interruption, respectively. Figures 1(b) and (d) are the strain rate versus strain curves of creep rupture and 1% plastic deformation, respectively, and figure 1(f) is the strain rate versus time curve at 20 h interruption. Under the same conditions, the creep rupture time of T13 was only half that of T6. The different rupture times resulted from the different creep rates, as shown in detail in figure 1(b). It is clear that the strain rate of T6 was smaller than that of T13 for most creep times. Only at the beginning and pre-rupture was the strain rate of T6 greater than that of T13, and then only for a short time.

A plastic deformation of 1% is a very important parameter for evaluating the creep property, which accounts for most of the creep time. As shown in figure 1(c), the 1% creep life of T6 was greater than that of T13, that is, ~180 and ~125 h, respectively. Though the 1% plastic deformation was minute, there were still numerous changes in the creep rate. As shown in figure 1(d), the creep rate increased at the beginning of the creep stage and maintained a constant value between ~0.02% to ~0.3%, and then decreased between ~0.3% to ~0.5% plastic deformation. With the continuing creep from 0.5% to 1%, the strain rate increased slowly. Although the change in strain rate is broadly in line with each other, there are two main differences that can be seen in figure 1(d). Between ~0.02% and 0.3%, the strain rate of T6 was clearly less than that of T13. After 0.5% plastic deformation, the strain rate of T6 increased more slowly than that of T13. These differences are a reflection of the micro-mechanism underlying this process, which will be discussed below.

The creep data for the first 20 h are plotted in detail in figures 1(e) and (f), which show very significant differences. It can be seen that a very slight creep of T6 is evident. By contrast, the plastic deformation of T13 is several times larger than that of T6. Furthermore, figure 1(f) provides more information about plastic deformation. The curve can be divided into three parts that include a decrease (0–2.5 h), an increase (2.5–10 h), and a second decrease (10–20 h). Unlike that for T13, the curve for T6 shows a very small fluctuation during the
first 20 h. Except for 18–20 h, the creep rate of T6 was always very small. The local negative creep rate was attributed to the dissolution and directional coarsening of the $\gamma'$ precipitate [37, 38].

### 3.2. Microstructure characteristics

Figures 2(a)–(e) shows the dislocation configuration of T13 and T6 after the creep test at 1100 °C/150 MPa. It can be seen that the square dislocation networks were established at the $\gamma/\gamma'$ interface at 1% plastic deformation (figures 1(c) and (e)). The dislocation network at 1% plastic deformation is the steady configuration; it remained stable at most times of creep. The dislocation spacing of the networks depends on the lattice misfit of the $\gamma$ and $\gamma'$ phases, and thus it can be used to calculate the misfit experimentally. As the dislocation networks were being woven during the early stages of creep, the investigation of dislocation development must be performed much earlier. As shown in figures 2(b) and (d), there is an obvious difference between the dislocation configurations of T13 and T6 following creeping for 20 h. For T13, the dislocation network was established completely even

![Figure 2. Interfacial dislocation networks in specimens documented by scanning electron microscopy (SEM). (a) Dislocation configuration of T13 superalloy crept at 1100 °C/150 MPa for 10 h; (b) 20 h; (c) 1% plastic deformation; (d) dislocation configuration of T6 superalloy crept under the same conditions for 20 h; (e) 1% plastic deformation.](image)

![Figure 3. Characteristics of $\gamma'$ rafting microstructure during the creep test. (a) 1% creep and (b) creep rupture for T13; (c) 1% creep and (d) creep rupture for T6. As the creep test proceeds, the $\gamma'$ raft is connected vertically to the next one crossing the matrix channel, as shown by the red dotted line in the photos.](image)
though the configuration was not very regular. However, the dislocation network of T6 had not formed at all after it crept for 20 h under the same conditions. Only a few rudiments of the dislocation network were documented, including position A in figure 2(d). In order to document the earlier configuration of the dislocation for T13, a 10 h interrupt creep was added. As shown in figure 2(a), the development stage of the dislocation network of T13 was still significantly posterior to T6 crept for 20 h. This suggests that the development of the dislocation network of T6 was slower than that of T13 under the same conditions.

Rafting is an important microstructure for nickel-based SC superalloys tested at high-temperature creep [39]. The rafting microstructures of T6 and T13 are shown in figures 3(a)–(d), which shows a very obvious and remarkable change from 1% plastic deformation to creep rupture: the separated $\gamma'$ rafts are connected to each other along the vertical direction. Because of the differences in the elemental content, the $\gamma'$ raft shows subtle differences in microstructure evolution. One can see that the vertical connection of the $\gamma'$ raft for T6 is more than that for T13, which is marked by red dotted lines. For the superalloy T6, the imperfect aging microstructure accelerated the connection of the $\gamma'$ rafting, which weakened the strength of the superalloy [40]. Though this connection of T6 was slightly greater than that of T13, the creep rupture life of T6 was significantly longer than that of T13.

3.3. Lattice misfit
It has been proved that the minimum creep rate is a function of the interfacial dislocation spacings [6, 7]. In this study, the dislocation spacings of the two superalloys were measured from the SEM microstructure. The measurement results were averaged from the experimental dislocation spacings of the superalloys. Because of
the specific orientation relationships [1], the [001] crystal plane in the SEM microstructure is easy to identify. The dislocation spacings for the T13 and T6 superalloys were 43.20 and 41.04 nm, respectively, in the [001] projected plane. As shown in figure 4, there is little difference in the dislocation spacings of T6 and T13. Because the dislocation lines extended on the \{111\} slip plane and $d = \frac{\sqrt{d_{001} \cdot \cos 45^\circ}}{2}$, the dislocation spacings on the \{111\} plane were 68.30 and 64.88 nm, respectively. The equation governing the relationship between dislocation spacing and misfit [6, 7] was used to calculate the misfits of the two superalloys. These results are also marked in figure 4. The lattice misfits of the two superalloys were very close to each other at 1100 °C, −0.38%, and −0.39%, respectively. The dispersion of the data for the dislocation spacing of T6 was slightly wider than that for T13.

### 3.4. Diffusivity

The sum of the contents of the refractory elements (W, Mo, Re, Ru, Ta) in the superalloys T6 and T13 were 24.5% and 24.6%, respectively, i.e., no significant difference in the total contents of the superalloys. Only when the content of specific elements was compared, the differences can be observed. As shown in figure 5(a), T13 had a greater quantity of Mo and Ta, while T6 had more Re and Ru. The atomic percentage of W in T6 was close to the percentage of W in T13. Equivalent diffusivity has been adopted as the parameter to predict the creep performance of newly designed superalloys [42]. Figure 5(b) shows the diffusion curves of elements in Ni, and table 2 lists their pre-exponential factors and activation energies. Figure 6(a) shows the partition behavior of the elements detected by energy-dispersive x-ray spectroscopy (EDS); (b) signal acquisition location of EDS. Both (a) and (b) show the results for T6; the behavior of T13 was similar.

| Element | $D_0$ [m$^2$ s$^{-1}$] | $Q_{\text{th, Ni}}$ [kJ mol$^{-1}$] | Element | $D_0$ [m$^2$ s$^{-1}$] | $Q_{\text{th, Ni}}$ [kJ mol$^{-1}$] |
|---------|----------------|-----------------|---------|----------------|----------------|
| Ni      | $2.26 \times 10^{-4}$ | 287             | W       | $8.0 \times 10^{-6}$ | 264          |
| Al      | $1.3 \times 10^{-4}$  | 256             | Mo      | $1.15 \times 10^{-4}$ | 281.3        |
| Ta      | $2.19 \times 10^{-3}$ | 251             | Re      | $8.2 \times 10^{-7}$  | 255          |
| Cr      | $5.2 \times 10^{-4}$  | 289             | Ru      | $2.48 \times 10^{-4}$ | 304.4        |
| Co      | $7.5 \times 10^{-5}$  | 285.1           | Hf      | $1.8 \times 10^{-4}$  | 287          |

Table 2. Pre-exponential factors and activation energies for diffusion of different atoms in Ni.

| Z | Element | Family | T6 | T13 |
|---|---------|--------|-----|-----|
|   | Atomic fraction (%) | Atomic error (%) | Atomic fraction (%) | Atomic error (%) |
| 13 | Al | K | 5.79 | 0.73 | 1.76 | 0.28 |
| 24 | Cr | K | 8.34 | 1.53 | 4.53 | 0.94 |
| 27 | Co | K | 21.04 | 3.85 | 18.61 | 3.86 |
| 28 | Ni | K | 52.73 | 9.66 | 68.82 | 14.27 |
| 42 | Mo | K | 0.95 | 0.19 | 2.73 | 0.6 |
| 44 | Ru | K | 5.33 | 1.04 | 0.85 | 0.2 |
| 72 | Hf | L | 0.01 | 0 | 0.1 | 0.02 |
| 73 | Ta | L | 0 | 0.01 | 0.57 | 0.11 |
| 74 | W  | L | 1.5 | 0.25 | 0.56 | 0.11 |
| 75 | Re | L | 4.29 | 0.72 | 1.46 | 0.28 |
| Totals | 100.00 | — | 100.00 | — |

Table 3. Compositions of the γ phases of the superalloys T6 and T13 as measured by EDS.
4. Analysis and discussion

4.1. Relationship between deformation and microstructure

There are close congruent relationships between creep performance and microstructure. Benefitting from the lower strain rate, the rupture life of T6 is nearly twice as long as that of T13, as shown in figures 1(a) and (b). In previous studies, the minimum creep rate was found to be proportional to the interfacial dislocation spacing [6, 7]. Nevertheless, this relationship was not found to be present in this study because of the comparable interfacial dislocation spacing. Therefore, there are other factors affecting creep performance in our current work.

Apart from the classic three-stage division of creep, 1% creep can also be divided into different stages, as shown in figures 1(b) and (d). The matrix microstructure reflects this division accordingly. Before creeping to ~0.02%, the strain rate increased significantly. During this stage, the dislocation nucleates and rapidly slips in the horizontal matrix channel. With the filling of dislocations in this channel, it becomes more and more difficult to force additional dislocations into the channel. Therefore, between a strain of 0.02%–0.4%, the strain rate was first maintained at a relatively high level and then decreased to a low level, which was related to the initial formation of the dislocation networks. Because the dislocation network exerted a very strong constraining force on the motion of the dislocations [6], the strain rate decreased as the dislocation network developed. Between
0.5% and 1% strain, the strain rate increased slowly, which indicated a slow evolution of microstructure, including dislocation networks and γ' rafts.

Though the curves show similar morphologies, some differences can also be seen. First, it took ~20 and ~40 h for superalloys T13 and T6, respectively, to creep to 0.3% strain due to the different strain rates. The microstructure of the creep interrupted at 20 h can help explain this phenomenon. As shown in figure 2, the development of the dislocation network of T6 was obviously less than that of T13. It is reasonable to conclude that the lower strain rate of T6 resulted from the low density of dislocations in the matrix. The factors affecting dislocation density will be discussed in section 4.2. Second, between ~0.4% and 1% strain, the increase in the strain rate for T6 was significantly slower than that for T13. The small difference in dislocation spacing (~2 nm) is not believed to have had a noticeable influence on creep behavior. The reason for this difference in strain rate will be discussed in section 4.3.

### 4.2. Effect of stacking fault energy on dislocation network development

The T6 and T13 superalloys displayed significant differences in the development of their dislocation networks. Misfit stress, refractory elements, stacking fault energy, and γ' precipitates are believed to be the four main factors affecting the movement of dislocations [1, 2, 28, 42, 43]. As a result of the purposeful alloy design used in the present investigation, the misfit stress in T6 was the same as the misfit stress in T13 and the fraction of γ' precipitates in T6 was the same as the fraction of γ' precipitates in T13. We have good reasons to exclude the effect of misfit stress and precipitate fraction on dislocation network development.

The refractory elements in the matrix channels can decrease the movement of dislocations [13], which is well known as ‘solution strengthening’ by researchers in the field of materials science. The elastic interaction between dislocations and solute atmosphere is an important mechanism of solution strengthening, which results from the differences in atom sizes of the refractory elements [2, 44]. Lattice misfit is the direct manifestation of this effect. For the superalloys T6 and T13, there was little or no lattice misfit at 1100 °C. That is to say, the elastic interaction in T6 did not have a different effect from the elastic interaction in T13 on dislocation motion during the formation stage of the dislocation network.

The cross-slipping of dislocations in the matrix channel is the main characteristic of high-temperature creep in the development stage of dislocation networks [45, 46]. Also, the matrix of nickel-based superalloys has a lower SF energy [26, 47, 48]. Thus, it is necessary to investigate the effects of SF energy. The SF in the FCC structure can be seen as a local lattice transformation from FCC to HCP [15]. The differences in the Gibbs free energies of the γ matrixes of the T6 and T13 superalloys were 0.04 × 10^4 J mol^-1 and 0.16 × 10^4 J mol^-1, respectively (table 2). Suppose the SF energy is contributed by the Gibbs free energy alone, the SF energies of T13 and T6 can be calculated as 82.52 and 20.63 mJ m^-2, respectively. The SF energy of Ni is most likely to lie in the range ~120–130 mJ m^-2 [49]. The SF energies reported in the literature for the γ phase of nickel-based SC superalloys lie in the range ~20–32 mJ m^-2, although the chemical compositions of these superalloys are significantly different [47, 48, 50]. Though the calculation of the SF energy is rough, the effect of SF energy on dislocation movement can also be obtained. The cross-slipping rate is controlled by the SF energy [51], which is an important mechanism for dislocation multiplication [52]. Due to the greater contents of Re and Ru, T6 had a lower SF energy than T13. We suggest that the lower SF energy limits cross-slipping propagation in the evolution of the dislocation network, which thereby results in the lower strain rate and slower development of the dislocation network for T6. This conclusion is robust, in the sense that other factors affecting dislocation slip and propagation have been reasonably excluded.

The energy barrier of Ni for the transformation from FCC to HCP is 0.28 × 10^4 J mol^-1. SF can be regarded as two layers of HCP structure in an FCC crystal [53]. Thus,

\[ \gamma_{\text{SF}} = \frac{4 \Delta G}{N_A \cdot \alpha} \]  (1)

where \( N_A \) is Avogadro’s constant and \( \alpha \) is the lattice constant. The lattice constant of Ni is 0.35 × 10^-8 m, and the calculated SF energy based on equation (1) is 142 mJ m^-2 at 1100 °C. The decrease in SF energy with decreasing temperature is ~5 mJ m^-2 per 100 °C; therefore, the calculated SF energy of Ni at room temperature is ~110 mJ m^-2, that is, slightly lower than the value of 120–130 mJ m^-2 reported in the literature [49]. Therefore, the method of calculating the SF energy based on the lattice transformation from FCC to HCP is feasible.

The spacing of two partial dislocations equals

\[ d = \frac{G \cdot b_1 \cdot b_2}{2\pi \gamma_{\text{SF}}} \]  (2)

where \( G \) is the shear modulus of the materials. For superalloy MAR-M002, the shear modulus of the γ matrix at 1100 °C is ~82 GPa [54]. Taking \( G = 82 \) GPa, the separated partial dislocation spacing is 18.3 nm for T6 and
4.57 nm for T13. The mathematical relationship between the cross-slipping frequency and the strain rate is difficult to establish, but some basic discussion is feasible. The cross-slipping frequency can be described by the Arrhenius expression

\[ f(T) = \nu \exp \left( -\frac{\Delta E_{\text{act}}}{k_B T} \right) \]  

(3)

where \( \nu \) is a pre-factor, \( k_B \) is the Boltzmann’s constant, \( T \) is temperature, and \( \Delta E_{\text{act}} \) is the activation energy. The activation energy is approximately twice the energy of the two partial dislocation constrictions [51], and it has been pointed out in the literature [55] that

\[ \Delta E_{\text{act}} \approx \frac{1}{5} G b^2 d \]  

(4)

where \( G \) is the shear modulus, \( b \) is the Burgess vector, and \( d \) is the spacing of the partial dislocations. As shown in figure 7(a), the activation energy increases linearly with dislocation spacing. The \( \nu \) in equation (3) can generally take the value of the Debye frequency, \( 10^{13} \text{ s}^{-1} \) [56]. Combining equation (3) and (4), the attempt frequency for cross-slipping can be calculated. It is obvious that the frequency of cross-slipping decreases exponentially with increasing dislocation spacing (figure 7(b)). The smaller the cross-slipping frequency, the lower the creep rate and dislocation density during the primary creep.

4.3. Key role of diffusivity on steady creep

Above \( \sim 0.5\% \) deformation, creep enters into the steady creep stage, with dislocation network formation and \( \gamma’ \) rafting being the two main factors for this transformation. This microstructure degenerated very slowly even at 1100 °C and 150 MPa. The cutting of the [010] super dislocation into the \( \gamma’ \) raft [57] and the topological inversion of the \( \gamma’ \) raft [58] were believed to be the rate-controlling process of steady creep. In this paper, as shown in figure 7(d), the increase in strain rate for T6 was significantly less than that for T13. The key reason for this difference will be discussed later in this section. The minimum strain rate is a function of the interfacial dislocation spacing [7]. A series of studies have shown that the effect of dislocation spacing is significantly greater than that of diffusivity [7, 59]. The interfacial dislocation spacings of the two superalloys studied here, however, are nearly equal at 1% plastic deformation. The difference is less than the error (figure 3). That is to say, in our investigation, the dislocation spacing had equivalent effects on the strain rate above 0.5% deformation.

Slow microstructure changes are the typical characteristics of the steady creep stage. This stability manifests itself in two ways: the stability of the dislocation networks and the stability of the \( \gamma’ \) raft. Only a very small number of dislocations can be documented in the \( \gamma’ \) phase, and dislocation networks have regular shapes in interrupted specimens. In this stage, the cross-slipping of interfacial dislocations is very inapparent because the dislocations had already filled the \( \gamma \) matrix and formed a steady network. The different SF energy of the \( \gamma \) matrix, therefore, should not be the cause of the difference in creep rate after 0.5% strain.

The lower strain rate of steady creep results from the diffusivity of the atoms. On the one hand, the [010] dislocation entering the \( \gamma’ \) raft is controlled by climbing, for which the rate of climb is given by
\[ v_c = \frac{D_v X_v}{b} \]  

where \( D_v \) is the diffusivity of a vacancy and \( X_v \) is the vacancy concentration \([39]\). The entry of dislocations into the \( \gamma' \) rafting causes damage to the microstructure and accelerates the topological inversion of the raft \([40]\).

Equation (5) suggests that the vacancy diffusion rate is important for dislocation climb. Because of the alloying of W, Mo, Re, and Ru, the diffusivity of vacancies is significantly lowered \([4, 6, 14, 24, 60–62]\) and the damage done by dislocations to the microstructure is reduced. On the other hand, the diffusivity also influences the migration of atoms in the matrix, which is the key step for \( \gamma' \) raft topological inversion \([63]\). The latter must be assisted by long-range atomic migration. Therefore, the slower the atom migration, the slower the raft degeneration. The equivalent diffusion coefficient has been used to estimate the performance of superalloys in the design of new materials \([42]\). In this study, the diffusivity of T6 was lower than that of T13, meaning that the unbroken rafting could be maintained for a longer time, corresponding to the slower increase in the strain rate above \( \sim 0.5\% \) plastic deformation (figure 1(d)).

5. Conclusions

Two nickel-based SC superalloys have been investigated using creep testing at 1100 °C and 150 MPa and SEM observations. The following conclusions can be drawn from the results of our study:

1. Although a 1% creep strain is very small, creep performance can also be divided into different stages according to the characteristics of the strain rate versus strain curves. The T6 and T13 superalloys showed similar changes as a function of creep, but there were still some differences, including microstructure and strain rate, which were considered to be consequences of the different stacking fault energies and diffusivities of the matrix.

2. SF energy plays a key role in the development of dislocation networks. The activation energy of the dislocation cross-slipping of T6 in the matrix was greater than that of T13, which decreased the frequency of cross-slipping, hindering the multiplication of dislocations in the primary creep.

3. The main effect of refractory elements on creep is to decrease the diffusion coefficient of the \( \gamma \) matrix, which decreases the creep rupture life by reducing the strain rate of steady creep. On the one hand, this reduces the penetration of dislocations into the \( \gamma' \) raft that destroys the microstructure and, on the other hand, it slows down the topological inversion of the \( \gamma' \) raft.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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