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Influence of Ru on creep behaviour and concentration distribution of Re-containing Ni-based single crystal superalloy at high temperature

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Abstract

The influence of Ru on the creep behaviour of a Re-containing Ni-based single crystal alloy at high temperature is investigated by creep capabilities measurement and microstructure evaluation. The creep life of the Re alloy at 1100 °C/137 MPa obtained is 164 h, after added 3.0%Ru, the creep lifetime of 4.5Re/3Ru alloy at the same conditions is 321 h. The deforming characteristic of the alloys during creep is dislocations gliding on $\gamma$ matrix and climbing over $\gamma'$ phase. And a few dislocations of shearing $\gamma'$ phase may cross-glide for forming Kear-Wilsdorf (KW) locks in Re-containing alloy. While more KW locks reserve in the $\gamma'$ phase of 4.5Re/3Ru alloy. The reason of 4.5Re/3Ru alloy displaying an outstanding creep properties is put down to more Re, W atoms being distributed in the $\gamma'/\gamma$ interface to delay dislocations shearing $\gamma'$ phase and depress the sliding and cross-sliding of dislocations. Wherein, the atoms Re can replace the Al atoms in $\gamma'$ phase, the stronger bonding force between Ru atoms and Re, W, Mo promotes much more ones dissolving into $\gamma'$ phase to delay the diffusion of them, this is deem as an important reason of the Re/Ru alloy displaying an outstanding creep resistance at high temperature.

1. Introduction

Microstructure of Ni-based single crystal alloys is composed of cubical $\gamma'$ phase embedded coherent in $\gamma$ matrix, and the alloys had been commonly applied in some critical fields due to the outstanding high temperature (HT) mechanical properties, for example, making the blade parts of the advanced aero-engines [1]. With the raise of the inlet temperature in the turbines of aero-engine, the temperature capacity of alloys is required to improve. The more the refractory atoms, for example Re, Ru, W, Mo and Ta, the higher temperature capacity of alloys, and the content of the ones in alloys increases to 20.7% of third generation alloys [2]. Particularly, the temperature capacity of alloys increases 30 °C and 60 °C when the concentration of Re enhances to 3% and 6%, respectively, which is a content feature of the advanced Ni-based alloys [3].

Adding Ru in the 4.5% Re single crystal alloys could depress the precipitating of TCP phase [4] for mending the temperature capacity substantially [5, 6]. At present, the fourth generations Ni-based alloys contain the 2 ~ 4%Ru and 5 ~ 6%Re [7], so that the alloys display an outstanding HT mechanical properties. And the microstructure and capabilities of Re/Ru-containing Ni-based alloys are widely studied [8]. Although there are numerous investigations about the impact of Re/Ru on the HT creep behaviour of Ni-based alloys [9–11], the reason of the Ru improves the HT creep behaviours of Re-containing alloys at is still unclear.

The Ru improving obvious the strength of $\gamma$ matrix is recognized as a reason for Ru improving the high temperature creep resistance of alloys [8]. It is considered that the Ru can not enhance the extent of solid solution strengthening, but the Ru can greatly enhance the solubility of Re and W in $\gamma'/\gamma''$ phases [9]. And the influence of Ru on the creep resistance of alloys depends not only on the role of itself, but also on the interaction between elements [12], and which exhibits close relevance with the content of Ru. Even though the high temperature creep behaviour of Ru-containing alloys had been bulletined [13], but the influence of Ru on the HT creep
The measured creep curves of Re-containing alloy in the steady state period of creep are acquired as around 0.012%/h and the creep lifetime is 164 h while that of Re/Ru-containing alloy are about 0.0066%/h, 321 h, respectively. These results demonstrated that the increasing extent of the lifetime obtains to 95.7%, indicating that the HT creep lifetime of Re-containing alloy would be significantly enhanced by the Ru.

The creep curves of Re/Ru-containing alloy with applying various stresses at 1070 °C ~ 1100 °C are expressed in figure 2, and the capabilities of alloy with applying various stresses at 137 MPa are expressed in figure 2(a). Here, the strain rates of the alloy in the steady state period of creep at 1070 °C, 1085 °C and 1100 °C are about 0.0066%/h, 0.012%/h and 0.014%/h, respectively, and the creep lifetime are 0.32 h, 164 h and 321 h, respectively. The strain rate of the alloy in the steady state period of creep at 1070 °C, 1085 °C and 1100 °C are significantly enhanced by the Ru.

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1100 °C are about 0.0022%/h, 0.0038%/h and 0.0066%/h, the lifespan of the alloy are recorded as 461 h, 386 h and 321 h, respectively, to exhibit a better creep property during creep.

Once applied stress at high temperature, the rapid multiplication of dislocations in γ matrix occurred, caused by the transient strains of alloys, so as to filling among the cubical γ′ precipitates, which corresponds to the initial stage of creep (including elastic strain). And with the prolongation of creep time, the denser dislocations are piled up in γ matrix to reduce the strain rate of alloy attributing to the accumulation and multiplication of dislocations. Meanwhile, the gliding and climbing of dislocations give rise to the recovery softening, under the action of heat activation, to release the stress concentration of local area. The creep of alloy entries steady state period once the balance of strain strengthening and recovery softening is acquired, the strain rate of alloy during creep could be indicated by Norton-Baily law [14].

\[
\dot{\varepsilon} = A \sigma^n \exp \left( -\frac{Q}{RT} \right)
\]  

Hereby, A being a constant, R representing gas constant, \( n \) – stress exponent, \( \sigma \) – applied stress, T – absolute temperature, \( Q \) – activation energy of creep. Based on the data in figure 2, Creep curves of Re-containing alloy are omitted, the relationship among strain rates, temperatures and stresses during steady state creep of alloys is indicated in figure 3.

According to the data in figure 3, the activation energies and stress exponents of the Re-containing and Re/Ru-containing alloys in the steady state period of creep are calculated to be \( Q_1 = 434.9 \text{ kJ mol}^{-1} \) and \( Q_2 = 509.8 \text{ kJ mol}^{-1} \), \( n_1 = 3.56 \) and \( n_2 = 5.68 \), separately. It’s obvious in the findings that the Re/Ru-containing alloy exhibits a better creep property, and the deforming characteristics of alloys in the steady state
period of creep being dislocations gliding in matrix and climbing over $\gamma'$ phase may be deduced based on the stress exponents.

3.2. Effect of Ru on components distribution

After crept until rupture at 1100 °C/137 MPa, the distribution of atoms in $\gamma$, $\gamma'$ phases of Re-containing and Re/Ru-containing alloys are acquired by 3D APT, to show in table 3 (the concentration distributions of alloys at heat treated state are omitted). Compared to the Re-containing alloy, the distribution of Al in $\gamma'/\gamma$ phases of Re/Ru-containing alloy increases from 4.43/1 to 4.84/1, the distribution of Re in $\gamma'/\gamma$ phases is changed to 1/5.02 from 1/7.64, the ratio of W in $\gamma'/\gamma$ phases is changed to 1/1.18 from 1/1.45, while the one of Co, Mo atoms in $\gamma'/\gamma$ phases does not significantly change.

As the comparison with the concentration distribution of alloy before creeping, after crept until rupture, the concentration of Cr, Al in $\gamma$, $\gamma'$ phases of Re/Ru alloy is reduced by a wide margin. Here, the content of Cr in $\gamma$ phase diminishes to 10.53 at% from 15.03 at%, the content of one in $\gamma'$ phase reduces from 2.04 at% to 1.87 at%, as indicated in table 3, which is put down to the reaction of Al, Cr with O on the surface of sample to be consumed during creep [14].

What’s more, the content of Re in $\gamma'$ phase of Re/Ru-containing alloy after crept is reduced to 0.33 at.% from 0.63 at.%, the distribution ratio of the one in $\gamma'/\gamma$ phases reduces from 1/5.02 to 1/10.75, while the concentration and ratio of other elements do not significantly change. After added Ru element, some of Ru atoms may substitute for Al atom in Ni$_3$Al phase [13], and the Ru atoms possess a stronger binding energy with Re, W and Mo atoms because of the hybridization and interaction of their d-electron layers [15], which may depress the diffusion of Re, W atoms around Ru during creep. Therefore, the Ru, Re, W in $\gamma$, $\gamma'$ phases keeps the stable concentration. In addition, the atom Re in $\gamma'$ phase is likely to be repelled, during creep, to dissolve in $\gamma$ matrix owing to its lower solubility in $\gamma'$ phase [16], this is one of reasons for reducing the concentration of Re in $\gamma'$ phase.

After crept until rupture at 1100 °C/137 MPa, the distributions of Al, Mo, W and Re in the area close to $\gamma'/\gamma$ interface of Re-containing and Re/Ru-containing alloys are indicated in figure 4. Wherein, the distributions of the elements in the Re alloy are labelled by the full lines, while the distributions of the elements in Re/Ru alloy are labelled by the dotted curves. Here, the concentration of Al at 0 point of the abscissa is C$_{Al}$ = 10 at.%, therefore, it

![Figure 3. Dependence of strain rates in the steady state period of creep on applied temperatures, stresses. (a) Strain rates & temperatures, (b) strain rates & stresses.](image)

| Alloys | Area   | Al  | Ta  | Cr  | Co  | Mo  | W   | Re  | Ru  | Total |
|-------|--------|-----|-----|-----|-----|-----|-----|-----|-----|-------|
| 4.5%Re| $\gamma$ phase | 4.89 | 0.51 | 10.53 | 10.79 | —   | 2.32 | 3.92 | 0   | 35.60 |
|       | $\gamma'$ phase | 16.53 | 4.51 | 1.87 | 3.62 | —   | 0.93 | 0.26 | 0   | 28.49 |
|       | Ratio  | 3.38/1 | 8.84/1 | 1/6.95 | 1/2.98 | 1/2.56 | 1/2.49 | 1/15.08 | —   | 1/1.25 |
| 4.5%Re/3.0%Ru| $\gamma$ phase | 3.91 | —   | 5.68 | 11.14 | —   | 1.91 | 3.55 | 3.30 | 33.63 |
|       | $\gamma'$ phase | 17.24 | —   | 0.77 | 3.70 | —   | 1.64 | 0.33 | 1.03 | 30.85 |
|       | Ratio  | 4.41/1 | 7.35/1 | 1/7.21 | 1/3.09 | 1/2.82 | 1/1.16 | 1/10.75 | 1/3.93 | 1/1.09 |
may be understood according to the concentration of Al that the phase in the left side of 0 point is defined as $\gamma'$, the one in the right side of 0 point is defined as $\gamma$.

Based on the definition of phase component [17], the concentrations of Al ($C_{Al} = 110\% \ C_{Al}^{\gamma}$ and $C_{Al} = 90\% \ C_{Al}^{\gamma'}$) are regarded as the component dividing lines of $\gamma$, $\gamma'$ rafts. Here, the content of Al in $\gamma$, $\gamma'$ rafts is defined as $\delta C_{Al} = \delta C_{Al}^{\gamma}$, respectively, as shown in table 3. The zone between two dividing lines is known as the transition one between $\gamma$ and $\gamma'$ phases, the zone between the full lines is determined as the transition one in 4.5%Re alloy, while the zone between the dotted lines is determined as the transition one in Re/Ru alloy, as indicated in figure 4.

The widths of the transition zone for Re-containing and Re/Ru-containing alloys at heat treated state are measured to be 3 nm (−2.0 to 1.0 nm) and 3.5 nm (−2.0 to 1.5 nm), respectively, the curves are omitted. After creep until rupture, the width of the transition zone of Re-containing and Re/Ru-containing alloys are measured to be about 4.45 nm (−1.25 to 3.2 nm) and 5.15 nm (−1.25 to 3.9 nm), as shown in figure 4. The results show that the width of the transition zone between $\gamma$ and $\gamma'$ phases of alloys increases with creep, which has a bearing on the morphology evolution of $\gamma/\gamma'$ phases among the creep progress. The morphology evolution of $\gamma/\gamma'$ phases increases their lattice parameters, as the creep goes on, to decrease their mismatch [18], as known as the dominant cause for increasing the transition zone width in alloys after creep until rupture.

On the one hand, more W, Re atoms are dispersed in the $\gamma$ phase near $\gamma/\gamma'$ interface of Re-containing and Re/Ru-containing alloys to form the high concentrations of them, as labelled by the arrows in figure 4. But the peak extent of W and Re concentration in the Re/Ru-containing alloy is smaller, which is related to some of Re, W atoms in $\gamma'$ phase being excluded for distributing in the $\gamma$ matrix near interface during creep.

As the creep goes on, the $\gamma/\gamma'$ interface is migrated toward $\gamma'$ phase to repel the Re, W atoms from $\gamma'$ phase for distributing in $\gamma$ matrix close interface to decrease the fraction of $\gamma'$ phase. On the other hand, it is difficult for the Re, W and Mo atoms to wide-range diffuse in $\gamma$ matrix owing to their big radius and lower coefficient of diffusion, consequently, the ones may enrich in the $\gamma$ matrix near interface. Moreover, the plused Ru atoms raise the content of W in $\gamma'$ phase to diminish the ratio of them in $\gamma/\gamma'$ phases, as indicated in table 3 and figure 4.

3.3. Effect of Ru on deformation mechanism

After heat treated, the microstructure of Re and Re/Ru alloys is comprised of $\gamma'$ and $\gamma$ phases, and the cubical $\gamma'$ phase is arrayed along (100) directions. In addition, some fine $\gamma'$ precipitates have been scattered in $\gamma$ matrix of Re/Ru-containing alloy, as labelled by the arrow in figure 5. The size of cubical $\gamma'$ phase is 0.4 $\mu$m approximately, that of $\gamma$ phase is around 0.08 $\mu$m, the fraction of $\gamma'$ phase in the alloy is 68% approximately.

The microstructure of Re-containing alloy after creep for various time at 1100 °C/137 MPa is indicated in figure 6, the way of applied stress is labelled by the arrows. After creep for 50 h, the $\gamma'$ phase in alloy transforms into rafted structure, as indicated in figure 6(a), dislocation networks can be distributed in rafted $\gamma/\gamma'$ interfaces due to their bigger misfits, as exhibited in the white zone, the magnified morphology is shown in the right upside of figure 6(a). During creep, a great amount of dislocations glid in $\gamma$ channel, as indicated in the zone A, the dislocations of gliding to $\gamma/\gamma'$ interface react with the networks to alter their gliding direction, for boosting the dislocations climbing over $\gamma'$ rafts. But no dislocation is detected in $\gamma'$ phase of alloy.

After creep for 164 h until rupture, the dislocations morphology in the Re-containing alloy is indicated in figure 6(b), the trace of dislocations $A_1$ and $A_2$ is parallel to the stress axis, while that of dislocation $B$ is parallel to

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**Figure 4.** Distributions of Al and W, Mo, Re in the zone near $\gamma/\gamma'$ interface of Re and Re/Ru alloys after crept until rupture.
Based on the dislocation invisible criteria \( \mathbf{b} \cdot \mathbf{g} = 0 \), Burgers vector of the dislocations \( A_1, A_2 \) is determined to be \( \mathbf{b}_{A1} = \mathbf{b}_{A2} = a[10\bar{1}] \), the line vectors of them are parallel to \( [002] = \mathbf{\mu}_A \). Consequently, the gliding plane of dislocations \( A_1, A_2 \) are determined to be the \( (010) \) plane owing to \( \mathbf{b}_{A1} \times \mathbf{\mu}_A \). In addition, the \( \mathbf{b}_B = a[110] \) is determined to be Burgers vector of the dislocation \( B \), and the gliding plane of the one is determined as \( \mathbf{b}_B \times \mathbf{\mu}_B = (\bar{1}11) \) owing to the line vector of \( B \) being \( \mathbf{\mu}_B = [022] \). The double-line contrast of dislocation \( A_2 \) is attributed to its decomposition for forming the partials plus anti-phase boundary (APB). The decomposition reaction of dislocation on the \( (010) \) plane can be expressed as follows:

\[
\mathbf{a}[10\bar{1}] \rightarrow \left( \frac{\mathbf{a}}{2} \right)[10\bar{1}] + \text{(APB)}_{(010)} + \left( \frac{\mathbf{a}}{2} \right)[10\bar{1}] \quad (2)
\]

The microstructures of the Re/Ru-containing alloy creeping for various times at 1100 °C/137 MPa are presented in figure 7 and arrows represents the direction of applied stress. Creeping for 80 h later, the \( \gamma' \) phase in alloy transforms into rafted structure, as shown in figure 7(a), for displaying the non-uniform size in thickness. The size of the rafted \( \gamma' \) phase in the upside of the photo is smaller of around 0.4 \( \mu \text{m} \), but the one in the middle of photo is around 0.8 \( \mu \text{m} \), the size of \( \gamma \) channel is 0.2–0.3 \( \mu \text{m} \). Moreover, some of fine \( \gamma \) phase are located within the thicker \( \gamma' \) rafts, as labelled by the white arrow in figure 7(a), suggesting that the coarsening of \( \gamma' \) rafts is concerned with the consolidation of adjacent rafted \( \gamma' \) phase, fine \( \gamma \) phase within them is gradually dissolved by diffusing. Some dislocations glide in \( \gamma \) phase during creep, the dislocations networks are located in the interfaces between \( \gamma' \) and \( \gamma \) phases, only a few dislocations shear into \( \gamma' \) phase, as labelled by the black arrow in figure 7(a). Consequently, it is deduced the deformation characteristic of alloy in the steady state of creep is dislocations sliding in \( \gamma \) phase and climbing over \( \gamma' \) strengthening one.

The morphology of alloy after creeping for 260 h at 1100 °C/137 MPa is exhibited in figure 7(b), exhibiting that the \( \gamma' \) phase maintains still rafts morphology, and the twisted extent of the one aggravates. Some dislocations glide still in \( \gamma \) phase, parts of ones have sheared \( \gamma' \) one. The traces of dislocations shearing \( \gamma' \) phase
are identified as [001] and [011] directions, which are parallel or around 45° angles referring to the stress axis, respectively, as labelled by the letters A, B and arrows on figure 7(b). Moreover, the regular dislocation networks are located in the $\gamma'$/$\gamma$ interfaces. After creeping for 321 h until fracture at the same conditions, the morphology in the zone apart from fracture is indicated in figure 7(c), where the $\gamma'$ rafts are twisted due to shearing of dislocations. Wherein, the traces of dislocations shearing $\gamma'$ rafts are perpendicular or parallel to stress axis, respectively, as labelled by C, D and arrows (figure 7(c)).

After the Re/Ru-containing alloy is crept until rupture at 1100 °C/137 MPa, the microstructures in the zone near fracture is presented in figure 8, great number of dislocations shearing $\gamma'$ rafts exhibit the parallel or vertical traces referring to stress axis, as labeled by the arrows. Thereinto, the double-lines contrast of dislocation F is attributed to its decomposition to transform into the partials plus APB. Based on the invisible criterion, the $b_E = a[10\bar{1}]$ is determined to be Burgers vector of dislocations E, the trace of the one is $\mu_E = [001]$, consequently, the gliding plane of dislocation E is determined to be $b_E \times \mu_E = (010)$. The $b_F = a[01\bar{1}]$ is determined to be Burgers vector of dislocations F, and the gliding plane of dislocation F is defined as $b_F \times \mu_F = (100)$ owing to its trace being parallel to $\mu_F = [010]$.

It is considered by analysis that the dislocations A, B, C, D, E and F during creep are firstly activated for gliding on the {111} plane of $\gamma$ matrix to pile up in $\gamma'$/$\gamma$ interface. And the stress concentration goes up with the amount of dislocations, which make the dislocations shearing into $\gamma'$ phase. Among them, the dislocations in $\gamma'$ phase is likely to be cross-glided from {111} to {100} planes for forming KW locks. And the dislocations on {100} plane may decompose, according to the formula (2), to transform into the morphology of partials plus APB. The configuration with non-planar core structure can depress the gliding and cross-gliding of dislocations to improve the creep resistance of alloy. However with the creep continuing, the dislocations in KW locks is

Figure 7. Microstructures of Re/Ru alloy creeping for various times at 1100 °C/137 MPa. (a) Crept for 80 h, (b) crept for 260 h, (c) crept for 321 h up to fracture.

Figure 8. Microstructures in the zone near fracture of Re/Ru alloy crept until rupture at 1100 °C/137 MPa.
likely to be released for re-gliding on \{111\} plane, as labelled by the letter B in figure 6. Hence, only a few KW locks are retained in the Re alloy after crept until rupture.

But for Re/Ru alloy after crept until rupture, the reason of more KW locks retaining in $\gamma'$ phase is put down to the interactivity of Ru and Re, W. The Ru may replace the Al atoms in $\gamma'$-Ni$_3$Al phase. The interactivity of the Ru and Re, W atoms [15] enhances their bonding energy, so that more Re, W atoms are reserved in $\gamma'$ precipitates to enhance alloying extent, which can depress the diffusion of elements and gliding of dislocations, so more KW locks are reserved in $\gamma'$ precipitates of Re/Ru alloy after crept until rupture at 1100°C/137 MPa. Hence, compared with Re alloy, the Re/Ru alloy exhibits a better creep property owing to more KW locks maintaining in $\gamma'$ phase to depress the dislocations gliding.

Accordingly, it’s acknowledged that the deforming feature of alloy in the later period of creep is the alternative operations of the initial/secondary gliding systems. During creep, the initial gliding system is firstly opened to contort the $\gamma'/\gamma$ phases, then the secondary gliding system is operated for shearing the initial gliding system [19], causing the micro-cracks appearing within the intersecting region of initial / secondary gliding systems [20]. As the creep continues, the micro-cracks are propagated, and the creep rupture of alloy occurs when the micro-cracks at different cross-section are gradually joined though the tearing edges [19]. This is the destroy and fracture characteristics of alloy during creep.

4. Discussion

The element Ta is $\gamma'$ former, adding Ta element may enhance the strenght and HT creep resistance of $\gamma'$ phase, hence the same content of Ta adds in the Re and Re/Ru alloys. But the various Cr contents of about 5.38 wt% and 2.78 wt% are, respectively, added in the Re and Re/Ru alloys, as indicated in table 1, which is related to the Cr self-reliance and interaction of Cr with other elements.

On the one hand, adding Cr may enhance the oxidation and corrosion resistance of nickel-based superalloys [21]. But the radius of Cr atom exhibits similarity with the one of Ni atom, consequently, added Cr atoms in nickel-based alloys has only the weaker solution strengthening effect, and can not significantly promote the HT creep capabilities of alloys [22]. On the other hand, when the elements Ru and Cr are, respectively, added to the nickel-based alloys, the rate of $\gamma'$ rafting during creep is accelerated to promote the regular dislocation networks dispersing in the interface of rafted $\gamma'/\gamma$ phases, which can improve the creep capability of alloy [23]. Furthermore, adding Ru to the 2.8 wt% Cr alloy, the interaction of Cr and Ru atoms may significantly improve the creep capability of alloy. However, adding Ru into the 5–6 wt% Cr alloy can promote the precipitation of TCP phase, which has the possibility to significantly reduce the creep capability of alloy [24]. Therefore, the lower Cr content shall be maintained in the Re/Ru alloy.

Owing to the interactivity of the atoms Ru and Re, W, more Re, Ru, W and Mo atoms dissolve in $\gamma'$, $\gamma'$ phases to enhance the content (c$_f$) of refractory elements and lattice aberrance, misfits (C$_o$) of the $\gamma'$ and $\gamma$ phases, which his likely to promote the creep resistance of alloy. Hence, the Re/Ru alloy exhibits a better creep capability at HT. Because the $\gamma'$-Ni$_3$Al with ordering structure is the main strengthening phase of alloy, the creep resistance of alloy grows with the radius (r) and volume fraction (f) of $\gamma'$ phase.

Furthermore, the dislocations in $\gamma'$ phase can be decomposed for transforming into the partials plus APBs in the late part of creep, for instance dislocations C and D in figure 7 and F in figure 8, which can produce the anti-phase boundary energy ($\eta_{APB}$) [25]. The APB energy ($\eta_{APB}$) can impede also the dislocations moving, therefore, the critical stress ($\Delta\tau$) of $\gamma'$ phase being sheared can be expressed as [26]:

$$\Delta\tau = \frac{A \cdot \mu \cdot (C_o) (ct \cdot r \cdot f \cdot \eta_{APB})^{1/2}}{b}$$

Here, A being a constant related to the amount of W, Re atoms in the zone of interface, $A = 3$ corresponding to the Re/Ru-containing alloy, and $A = 1$ corresponding to the Re-free or Re-containing alloys, $\mu$ being the shearing modulus, $b$ being Burgers vectors, $T$ being the line tensor. The formula (3) expresses the critical shearing stress of alloy during creep grows with the content (c$_f$) of the refractory atoms and the fraction (f), size (r) of $\gamma'$ phase.

Besides, the dislocations in $\gamma'$ phase can decompose on \{100\} planes for forming the morphology of partials plus APBs, as shown in figures 6(b), 7(c) and 8, and schematically shown in figure 9. The configuration of KW locks plus APB can effectively depress also the cross-sliding and sliding of dislocations, hence, the creep properties of alloy increases with the amount of KW locking.

Because the higher content (c$_f$) of refractory atoms, big size (r) and fraction (f) of $\gamma'$ phase, more KW locking are reserved in the Re/Ru alloy, and more refractory atoms are distributed in the zone near $\gamma'/\gamma$ interfaces, the alloy at high temperature performs a better creep properties.
5. Conclusion

(1) Compared with Re alloy, the creep lives of Re/Ru alloy at 1100 °C/137 MPa enhances from 164 h to 321 h. The deformation mechanisms of alloys in the late period of creep are dislocations shearing γ′ phase. And the dislocations of shearing γ′ rafts can cross-glide for forming KW locks.

(2) The cause of the Re/Ru alloy exhibiting an excellent creep property is ascribed to more W, Re atoms distributing in γ′ phase to reserve the KW locks which can depress the sliding and cross-sliding of dislocations at high temperature.

(3) The interactivity of Ru and W, Re make more refractory atoms dissolving in γ′ phase to increase its alloying extent. And more atoms W, Re distributing in γ phase near interface can depress dislocations movement and diffusion of other elements, which is one of caused for more KW locks reserving in γ′ rafts of the Re/Ru alloy at 1100 °C.

(4) In the late period of creep, the sliding of dislocations has been alternately operated to twist the γ′/γ rafts, which could promote the initiation of cracks along the γ′/γ interfaces. As the creep continues, the cracks have been propagated along with γ′/γ interfaces until rupture, which is the destroy and fracture feature of alloys during high temperature creep.

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