Creep Characteristic of NiAl-9Mo Eutectic Alloy

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The creep behavior of NiAl-9Mo eutectic alloy has been studied in an effort to understand the characteristic that the alloy exhibits of the low stress exponent of 4.75 and the high apparent activation energy of 410 kJ/mol during the steady-state creep of under the test conditions. The material exhibits threshold behavior with a threshold stress of 17.9 MPa. TEM observation reveals that the creep deformation is mainly governed by dislocation climb in NiAl matrix phase. Large recoverable strain is observed after unloading under the transient creep and increases with the increase of stress or temperature. It exhibits an activation energy 232 kJ/mol, suggesting the processes is controlled by the flow behavior of the matrix phase. The apparent creep activation energy under constant stress is a combination of the creep flow activation energy of the matrix and the activation energy for the damage process at the phase interface.

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1. Introduction

Creep property is an important factor that controls the serving lifetime of structural materials at high temperatures. In order to improve the elevated temperature creep strength, various particle hardened alloys or fiber reinforced composites are being developed. The creep behavior of these alloys is dependent on the creep properties of each phase, the interfacial interactions among each phase, the volume fraction and morphology of each phase.

NiAl based intermetallics are attractive candidates for high temperature structural applications because of their relatively low density, high melting point, good thermal conductivity and excellent oxidation resistance. However, poor damage tolerance at ambient temperatures and inadequate strength at high temperatures have prevented it from commercial applications. Fabricating eutectic alloys with refractory elements is an effective way to overcome the limitation of NiAl. NiAl-based eutectic alloys such as NiAl-Cr, NiAl-Mo, NiAl-Cr(Mo) have attracted a great interest owing to the combination of reasonable room-temperature toughness and high temperature creep strength. Fracture toughness of polycrystalline NiAl is 5~6 MPa√m, while that of NiAl-Cr/Mo eutectic alloys is 15~20 MPa√m. The extrinsic toughening mechanisms like crack trapping, crack deflection, interface debonding and crack bridging are responsible for the improvement in toughness. The creep-resistant secondary phases including refractory metal and Laves phase (NiAlX, X=Ta, Nb) provide the improvement of high temperature strength for the eutectic systems. Moreover, the NiAl-refractory metal alloys exhibit a pseudo-binary eutectic equilibrium and good chemical compatibility even up to the melting point, which makes the composite system promising for industrial use.

The creep strength of NiAl-based eutectic alloys was compared with that of binary single-crystal in terms of stress required to produce a creep rate of $10^{-6}$ s⁻¹ at 1300K by Johnson et al. Their investigation was performed with compression test. The creep parameters of NiAl-based eutectic alloys are summarized in Table 1. For comparison, those of single phase NiAl are also listed. It can be seen that the stress exponent of the eutectic alloy systems is the same as that of NiAl single phase and the apparent activation energy is much higher than that of NiAl single phase. It is wellknown that the second phase strengthened alloys show the high stress exponent in the range of 6~40 and high apparent creep activation energy, such as Al matrix alloys reinforced with Al₂O₃ fibers, SiC whiskers and SiC.

<table>
<thead>
<tr>
<th>Alloys (at%)</th>
<th>Temperature, K</th>
<th>Stress exponent</th>
<th>Apparent activation energy, kJ/mol</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>DSNiAl-28Cr-6Mo</td>
<td>1200~1400</td>
<td>6.38</td>
<td>457</td>
<td>4</td>
</tr>
<tr>
<td>HIPNiAl-9Mo</td>
<td>1123~1223</td>
<td>4.75</td>
<td>410</td>
<td>8</td>
</tr>
<tr>
<td>DSNiAl-16.5Nb</td>
<td>1200~1300</td>
<td>4.34</td>
<td>420</td>
<td>9</td>
</tr>
<tr>
<td>DSNiAl-14.5Ta</td>
<td>1200~1400</td>
<td>4.76</td>
<td>505</td>
<td>10</td>
</tr>
<tr>
<td>DSNiAl-5Nb-10Mo</td>
<td>1200~1400</td>
<td>4.66</td>
<td>479</td>
<td>11</td>
</tr>
<tr>
<td>Ni-(44-50.6)Al</td>
<td>1073~1318</td>
<td>4.42~6.35</td>
<td>254~334</td>
<td>12</td>
</tr>
</tbody>
</table>

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particles,\textsuperscript{15}\hspace{0mm} oxide dispersion strengthened superalloy\textsuperscript{16,17} and nickel-base superalloy containing a hard cubic $\gamma'$ precipitates. The NiAl-based eutectic alloys are reinforced with refractory metal and hard-brittle Laves phase. It is not clear why the stress exponent for the alloys is so low as that for the simple alloy and the apparent activation energy for the alloys is much higher than that of single phase NiAl.

In the present study, we will analyze the relationship among the steady-state creep rate and the applied stress as well as the temperature in NiAl-eutectic alloys by tensile creep test. On the basis of TEM observations, calculation of the threshold stress and research on transient creep behavior, we will provide an insight into the creep feature of low stress exponent and high apparent creep activation energy for the NiAl-based eutectic alloys at high temperature.

2. Experimental Procedures

The Ni-45.5Al-9Mo(at\%) alloy used in this study was prepared by induction melting in vacuum and the purities of starting materials are 99.999%Al, 99.99%Ni, 99.9%Mo. The cast rods were processed by hot isostatically pressed (HIP) at 1300$^\circ$C/100 MPa for 2 hours. Creep specimens with a gauge section of $2 \text{ mm} \times 2.5 \text{ mm} \times 16 \text{ mm}$ were electro-discharge machined from the HIPed rods. Prior to testing, all samples were ground with 1000-grit SiC paper. The tensile creep tests were performed in a RCL-3 machine in a temperature range from 850 to 950$^\circ$C with a constant tensile stress range from 50 to 100 MPa in air. The test temperature was monitored with two thermocouples touching the specimen gauge section and held constant within $\pm 1^\circ$C for each individual test. A high sensitivity extensometer with a resolution of $5 \times 10^{-5}$ was attached to the gauge section of the specimens to measure displacement. Most of the creep tests were run to rupture, whereas some tests were interrupted for TEM observation.

The microstructures were examined using scanning electron microscopy (SEM) and transmission electron microscopy (TEM). SEM samples were etched in a solution of HNO$_3$:CH$_3$COOH = 1:10 (volume). TEM samples were machined mechanically to a thickness of $30 \mu$m and then thinned by ion milling. TEM investigation was conducted in a Philips TEM420 analytic electron microscope operating at 120 KV.

3. Results and Discussion

3.1 Microstructure

Figure 1 shows a SEM image of the NiAl-9Mo eutectic alloy. The microstructure typically consists of fine eutectic colonies. At the center, the Mo phase of small ovalshaped pellets with 0.2$\sim$2 $\mu$m indiameter is uniformly distributed in the NiAl matrix. These small pellets almost take up the main space in the grain cell. At the cell boundary, the Mo phase is fibrous. The length of these Mo fibers is in the range of 2$\sim$10 $\mu$m.

The Mo phase volume fraction of 12% was determined by area fraction measurements of metallographic sections. The composition of phases in atomic percent was determined by energy dispersive spectroscopy (EDS): NiAl phase Ni50.2, Al49.8, Mo0, and Mo phase: Ni4.2, Al8.5, Mo87.3.

3.2 Steady-state creep

Constant stress tensile creep tests were performed in air at temperatures of 850, 900 and 950$^\circ$C, with stresses between 50 and 100 MPa. These tests were run until the samples failure. Data from these creep tests are summarized in Table 2.

Table 2 Creep data of NiAl-9Mo eutectic alloy at the different temperatures.

<table>
<thead>
<tr>
<th>Temperature $^\circ$C</th>
<th>Applied stress MPa</th>
<th>Steady-state creep rate, $\text{s}^{-1}$</th>
<th>Rupture time h</th>
<th>Fracture elongation %</th>
</tr>
</thead>
<tbody>
<tr>
<td>850</td>
<td>60</td>
<td>$1.74 \times 10^{-7}$</td>
<td>314</td>
<td>40.6</td>
</tr>
<tr>
<td>900</td>
<td>50</td>
<td>$3.2 \times 10^{-7}$</td>
<td>203.4</td>
<td>42.3</td>
</tr>
<tr>
<td>950</td>
<td>50</td>
<td>$4.2 \times 10^{-6}$</td>
<td>14.5</td>
<td>47.5</td>
</tr>
<tr>
<td>100</td>
<td>60</td>
<td>$9.96 \times 10^{-6}$</td>
<td>4.5</td>
<td>51.9</td>
</tr>
<tr>
<td>100</td>
<td>50</td>
<td>$1.61 \times 10^{-6}$</td>
<td>42.8</td>
<td>42.8</td>
</tr>
<tr>
<td>60</td>
<td>$6.6 \times 10^{-6}$</td>
<td>13.2</td>
<td>50.3</td>
<td></td>
</tr>
<tr>
<td>80</td>
<td>$2.5 \times 10^{-5}$</td>
<td>5.5</td>
<td>41.4</td>
<td></td>
</tr>
</tbody>
</table>

The creep curves exhibit very short inverse primary stage which is followed by well-defined and long steady-state creep stage. During the primary creep stage, the creep rate decreases. Then, it remains constant to a strain of 10$\sim$15%. The specimens undergo extensive plastic deformation before fracture during the tertiary creep stage.

For NiAl-based intermetallics, the steady-state creep rate $\dot{\varepsilon}$ and constant applied stress $\sigma$ and temperature $T$ are fitted to the Dorn Equation,

$$\dot{\varepsilon} = A\sigma^n e^{-\frac{Q_{apr}}{RT}}$$  \hspace{0mm} (1)
where $A$ is a constant, $R$ is the universal gas constant, $n$ is the apparent stress exponent and $Q_{app}$ is the apparent activation energy for creep. The dependence of $\ln \frac{\varepsilon}{C_0} - \ln \frac{\varepsilon}{C_27}$ is shown in Fig. 3 at the temperatures of 850 and 950°C. From Fig. 3, the average value of power law stress exponent $n$ is determined to be 4.75. The dependence of the steady-state creep rate on temperature at constant applied stress is plotted in Fig. 4. Accordingly, the apparent creep activation energy $Q_{app}$ is calculated to be about 410 kJ/mol.

The stress exponent $n$ is believed to provide an insight into the creep mechanisms. The values of 1, 3, 5 are considered to represent the processes of grain boundary diffusion controlled creep (Coble creep),\(^{18}\) viscous glide of mobile dislocation (Class I),\(^{19}\) high temperature dislocation climb controlled creep due to lattice self-diffusion (Class II),\(^{20}\) respectively. The stress exponents for the creep deformation of binary NiAl reported by a number of authors scatter between 1 and 10.\(^{2,121}\) The origin of this wide scatter is not clear although the value of $n$ is known to be dependent on the temperature, stress and processing history. Most of $n$ values reported cluster between 5 and 7, and TEM observations of crept NiAl appear to lead to the conclusion that the creep behavior of binary NiAl is of Class type.\(^{13,22,23}\) The values of $n$ obtained in the present work for NiAl-9Mo alloy was approximately 4.75, which agrees with that for binary NiAl. It can be deduced that the creep rate of the alloy is governed by the dislocation climb in the NiAl phase. The TEM observation of creep specimen confirms the inference. Figure 5(a) shows the dislocation structure in NiAl and Mo phases. The dislocation density in NiAl matrix phase is significantly higher than that in Mo phase. Moreover, the Mo phase of only 12 volume percent is discontinuously distributed in the matrix. The strength of Mo at an elevated temperature is much higher than that of NiAl ($E_{Mo} = 300$ GPa, $E_{NiAl} = 151$ GPa, at 900°C).\(^{1,24}\) These are enough to indicate that the creep process is accomplished mainly via the deformation in NiAl phase. The subgrain boundaries were also observed in NiAl phase from the Fig. 5(b). This means that dislocation climb takes place in the matrix NiAl phase during the creep process.

### 3.3 Transient creep test

The experimental procedure for transient creep is described in Fig. 6. When the temperature reaches the prescribed value, an initial stress is applied. After 5~10 min, the stress is reduced to $\sigma_2$. Approximately 10% of the load remains on the
The sample is allowed to relax for a period of time (20 min) before the major stress is reapplied. It can be seen that the specimen shows a large recoverable strain ($\varepsilon_r$) even though under the tension stress after unloading the major stress. The recoverable anelastic strain response does not stop; rather, the material continues to contract after the stress is removed, albeit at ever slower rates. Most of the recoverable strain is established during the initial part of the deformation. The magnitude of the recoverable deformation increases only slightly during the following course of the transient test.

Figure 7 shows the relationship among the recoverable strain, unloaded stress and temperature. In this plot, the magnitude of the strain recovered at 10 min after unloading is plotted against the magnitude of the stress removed. The magnitude of the recoverable strain increases with the increase of stress or temperature. Therefore, the anelastic backflow process is temperature dependent, with deformation occurring faster at the higher temperature. Furthermore, the process may be controlled by thermally activated flow in the matrix. So, the activation energy for this process is measured by determining backflow strain response to a change in temperature. Temperature change test on the alloy is shown in Fig. 8. The furnace has a thermal time constant of approximately 60 s for a 50°C temperature change, so that within 5 min, the sample has equilibrated at the new temperature. Therefore, the strain-time response at 5 min after the temperature change reflects the backflow strain rate for the new temperature. Backflow strain rates are expected to be larger in the time of the new thermal equilibrium, but are masked by the much larger positive thermal expansion.

Fig. 5 TEM observation of specimen crept for 12 hours under the condition of 900°C/60 MPa (a) the comparison of dislocation density in the matrix phase NiAl and reinforcing phase Mo (b) the subgrain boundaries in NiAl phase. The applied stress axis is vertical in the photos.

Fig. 6 Load/unload transient creep test conducted at 900°C, $\sigma_1$ is initially applied, stress is then reduced to $\sigma_2$ for a period time, $\sigma_1$ is reapplied until steady state is reached; stress is reduced once again to $\sigma_2$. 

-5 0 5 10 15 20 25 30 35 40
-0.000 -0.005 -0.010 -0.015 -0.020 -0.025

T=900°C
$\sigma_1=100$MPa, $\sigma_2=10$MPa

creep strain

creep strain

Time (min)

$\varepsilon_r$

$\sigma_1$

$\sigma_2$

$\sigma_1$

$\sigma_2$

$\sigma_1$

$\sigma_2$

$\sigma_1$

$\sigma_2$

$\sigma_1$

$\sigma_2$

$\sigma_1$

$\sigma_2$

$\sigma_1$

$\sigma_2$

$\sigma_1$

$\sigma_2$
strains. So, the backflow strain rate associated with the new temperature is calculated from the slope of the fitted curve extrapolated back to the time of the temperature change. The activation energy \( Q \) for the backflow process is calculated based on the pre-temperature-change strain rate and post-temperature-change maximum strain rate through the Arrhenius equation:

\[
\dot{\varepsilon} = A e^{-Q/RT} 
\]

The resulted value of \( Q \) is 232 kJ/mol.

4. General Discussion

4.1 Stress exponent and threshold stress

For metals and alloys at high temperature, the dependence of steady-state creep rate on constant applied stress and temperature can be usually described as Dorn eq. (1). For the simple alloys, the values of the apparent stress exponent is between 3 and 5, and the apparent activation energy for creep is close to that for volume self diffusion (\( Q_{sd} \)). For the second phase strengthened alloys, however, much high values of \( n = 6-40 \) are found, and the compatible values of \( Q_{ap} \) are also 1.5 to 3 times the value of \( Q_{sd} \). Some rationalization of these differences can be obtained by expressing the eq. (1) in terms of effect stress rather than the applied stress:

\[
\dot{\varepsilon} = A' (\sigma - \sigma_{th})^n e^{Q/RT} 
\]

where \( A' \) is a constant, \( \sigma_{th} \) is the threshold stress, \( n_o \) is the true stress exponent, \( Q \) is the true creep activation energy. The subtraction of \( \sigma_{th} \) from the applied stress results in the effective driving stress acting on the mobile dislocation during creep. \( \sigma_{th} \) can be estimated through iterative back calculations using isothermal creep data and eq. (3), provided \( n_o \) is known. A study shows that in most cases creep should be rate controlled by the recovery of dislocation networks formed during creep, \(^{17}\) in which \( n_o \) takes the value of 4. Indeed, experiment results do show that \( n_o \) is about 4, \(^{25-28}\) In the present study, \( n_o \) is taken as 4 because of the subgrain boundary formed during the creep test, and the eq. (3) can be written as:

\[
\dot{\varepsilon}^{1/n_o} = (A' e^{-Q/RT})^{1/n_o} (\sigma - \sigma_{th}) 
\]

The tested data is plotted in Fig. 9. Accordingly, \( \sigma_{th} \) is calculated to be 16.0, 20.0, 17.6 MPa at 850, 900, 950 °C, respectively. The average value is 17.9 MPa. Compared with the relative values of threshold stress in γ' strengthened nickel-based superalloys and oxide dispersion strengthened superalloys, \(^{17,25-28}\) as shown in Table 3, the alloy shows low threshold stresses.

The magnitude of threshold stress is determined by the dislocation motion in matrix phase being hindered by the second phase. \(^{17,25-28}\) The kinetics of dislocation climb over a particle is studied theoretically in the alloys reinforced with hard particles of low volume fraction by Rosler \(^{29,30}\) et al. They put forward two kinds of models for dislocation climb over particles: local climb and general climb. Local climb is an extremely unstable process. That is to say, the sharp bend in the dislocation line can be rapidly relaxed by diffusion, leading to a more general climb. Localization increases with increasing stress. Because fewer vacancies are required to support a more localized climb process, a natural power-law dependence of the dislocation velocity on applied stress, with \( n = 3 \) to \( 4 \), is obtained. In this case, the threshold stress is low. When dislocation generally climbs over hard particle, the particle exerts an attractive force on the dislocations. The attraction results in that the creep rate is extremely sensitive
to the applied stress, *i.e.* the alloy shows the high stress exponent and threshold stress. In the present study, the NiAl-9Mo alloy shows the low stress exponent and the small threshold stress. The interaction between the dislocation and the second phase may be explained by the local climb theory. Certainly, further work is needed to evidence.

### 4.2 Creep activation energy

It has been known that the creep process of the NiAl-9Mo alloy is controlled by the dislocation climb in the NiAl matrix phase. However, it is surprising that the apparent creep activation energy (410 kJ/mol) is much higher than that of Ni self-diffusion activation energy in NiAl (220–300 kJ/mol) and creep activation energy of binary NiAl (250–320 kJ/mol). Several attempts have been made to explain the reason for the high creep activation energy. A more accurate description of the activation energy can be found by compensating the apparent activation energy for creep of the alloy for temperature dependence of the elastic modulus by using the equation:

$$Q_c = Q_{\text{apparent}} - nR \frac{d \ln E}{dT}$$

Using the average value of $n = 4.75$ for the NiAl-9Mo alloy and the temperature dependence of the composite modulus (which was found to be the same as the temperature dependence of the matrix modulus), the true activation energy for creep of the composite is found to be $Q_c = 401$ kJ/mol. While this compensated activation energy is slightly less than the apparent activation energy quoted above, it is still significantly higher than the activation energy for creep of NiAl matrix. To probe the reason that the alloy shows the high apparent activation energy, stress change-strain transient creep tests were conducted on the alloy.

The material showed large recoverable strain after unloading the major stress in the transient creep test. When the material is deformed under the initial stress, the second phase and matrix undertake the elastic and plastic strain all together. The elevated temperature strength of Mo is much higher than that of NiAl ($E_{\text{Mo}} = 300$ GPa, $E_{\text{NiAl}} = 151$ GPa, at 900°C). Moreover, the Mo phase of only 12 volume percent is discontinuously distributed in the matrix. The size and morphology of the Mo phase in the grain cell remain unchanged after the sample crept. The strain of the second phase is from elastic deformation. The strain of matrix is a sum of the plastic and elastic deformation, in which the proportion of plastic strain is relatively large. After the alloy is unloaded, the matrix relapses from the elastic strain, while the second phase still remains elastic tension. Therefore, the elastic strain energy stored in the second phase provides a driving force for the contraction of the alloy. During the recover, the second phase contracts and compresses the matrix material in opposite direction of the applied load. Accordingly, the material shows the recoverable strain macroscopically. With the progress of recoverable deformation, the strain difference between the matrix and the second phase diminishes gradually, so the recoverable strain rate decreases.

The activation energy for the recoverable strain is 232 kJ/mol, which agrees with the creep activation energy for binary NiAl and Ni self-diffusion activation energy in NiAl. The close agreement suggests that the flow process in the alloy is controlled by the flow behavior of the matrix material. This further indicates that the high apparent activation energy for creep of the material under constant stress is controlled by other mechanisms, not directly controlled by the flow behavior of the matrix material.

The high apparent creep activation energy for Al/Al₂O₃ composite is discussed by T. L. Dragone et al. They concluded that the observed strain rate is not true steady state rate during the constant stress, but a minimum strain rate. The strain rate minimum is thought to be the result of a balance between a decreasing strain rate caused by load transfer to a continuous fiber network and an increasing strain rate caused by the accumulation of damage in the composite. Damage in the alloy may take the form of fracture of fiber network, void formation and debonding of the fiber-matrix interface. The minimum strain rate established when either of the two contributing processes results in strain rates that are equal to each other. The apparent activation energy determined by using minimum creep rates is a combination of both the activation energy for creep flow of the unreinforced matrix and the activation energy for the damage processes (though not necessarily the numerical sum). The theory can also explain that the activation energy for backflow in the study during transient creep is equal to that for flow in the matrix but not equal to the apparent activation energy for creep under constant stress. In the unload condition, the stress is not large enough to promote the formation and growth of voids and the separation of the second-phase-matrix interface.

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### Table 3 Threshold stress of the strengthened alloys

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Temperature range °C</th>
<th>Applied stress range MPa</th>
<th>Apparent stress exponent $n$</th>
<th>Threshold stress $\sigma_t$ MPa</th>
<th>$\sigma_t/\sigma$ average</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\gamma'$ strengthened superalloy</td>
<td>746–788</td>
<td>448–517</td>
<td>$7.64 \pm 0.03$</td>
<td>245</td>
<td>0.54</td>
<td>17</td>
</tr>
<tr>
<td>ODS solid solution superalloy</td>
<td>957–982</td>
<td>97–119</td>
<td>$6.75 \pm 0.69$</td>
<td>45</td>
<td>0.42</td>
<td>17</td>
</tr>
<tr>
<td>$\gamma'$ strengthened superalloy</td>
<td>941–996</td>
<td>190–258</td>
<td>$20.3 \pm 1.3$</td>
<td>169</td>
<td>0.75</td>
<td>17</td>
</tr>
<tr>
<td>NiAl-9Mo</td>
<td>830–900</td>
<td>280–420</td>
<td>$24.5 \pm 3.1$</td>
<td>96</td>
<td>0.83</td>
<td>28</td>
</tr>
<tr>
<td></td>
<td>850–950</td>
<td>50–100</td>
<td>$4.75 \pm 0.25$</td>
<td>17.9</td>
<td>0.28</td>
<td>This study</td>
</tr>
</tbody>
</table>
During the creep of constant stress, the applied stress causes the damage of the phase interface as shown in Fig. 10. Figure 10 is the appearance of the fracture surface for the creep sample of NiAl-9Mo alloy. Dimpling is apparent on the fracture surface. Most of the dimples are associated with the Mo phase, suggesting that the voids form at the phase interface and subsequently grow and coalescence. Therefore, the apparent creep activation energy under constant stress is a combination of the creep flow activation energy of the matrix and the activation energy for the damage process at the interface between the NiAl matrix phase and reinforcing phase Mo. This theory maybe also help to explain the high creep activation energy for some alloys reinforced with the second phase.

5. Conclusions

(1) The creep curves exhibit a very short primary, long steady-state creep stage and high fracture strain in the temperatures range of 850–950°C and constant tensile stresses from 50 to 100 MPa.

(2) The measured power law stress exponent for steady-state creep is determined to be 4.75 and the apparent creep activation energy is calculated as 232 kJ/mol. The steady-state creep deformation is governed by dislocation climb in NiAl matrix phase.

(3) The material exhibits threshold behavior, with the low threshold stress of about 17.9 MPa.

(4) The material shows large recoverable strain after unloading the major stress during the transient creep. The magnitude of the recoverable strain increases with the increase of stress or temperature. The activation energy of recoverable flow process is determined to be 232 kJ/mol. The process is controlled by the flow behavior of the matrix material.

(5) The apparent creep activation energy of the alloy under constant stress is a combination of the creep flow activation energy of the matrix and the activation energy for the damage process at the phase interface.

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