Creep Behavior and Threshold Stress of an Al-Al₂O₃-Zr Alloy*

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Creep tests and high temperature tensile tests were carried out on the coarse-grained dispersion-hardened Al-Al₂O₃-Zr alloy which had an anisotropy of strength due to the recrystallization texture. Three types of specimens were used for the tests, each of which had a specimen axis orientation of 0, 45 or 90 degrees against the rolling direction, respectively. Relations between the critical stress for creep deformation and the proof stress were examined.

The existence of critical stress for creep deformation, σ_c , was recognized at temperatures from 573 to 773 K. The specimens of 0 degree orientation had the highest critical stress and creep strength, and the specimens of 90-degree orientation had the lowest ones. It was shown that the creep deformation could be explained as being controlled by the self-diffusion of Al when it was taken into account that σ_c decreases with the rise of temperature. The temperature dependence of σ_c was thought to be due to the interaction between dislocations and impurity atoms which segregated on the interface between dispersed particles and the matrix. Ratios of σ_c to the 0.1 and 0.01% proof stresses at room temperature, which were thought to be a criterion of the Orowan stress, were in the range from 0.5 to 0.8. Each ratio was nearly constant at the same temperature independent of the axis orientation. The ratio decreased gradually with the rise of temperature, which showed that the interaction between the dislocations and the dispersed particles changes with the rise of temperature.

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I. Introduction

The creep behavior of dispersion hardened alloys shows some prominent characteristics which are different from other metals and alloys. The stress exponent of dislocation creep, which is about eight or more, is usually larger than that of pure metals and solid solutions⁽¹⁾. The creep rate decreases rapidly with decreasing stress below a stress, showing the existence of the threshold stress for the creep deformation⁽²⁾. Though it is difficult to measure the threshold stress explicitly, the stress is considered to be related to the critical stress for the beginning of plastic deformation, or the yielding stress associated with dispersed particles.

The yielding of the matrix containing dispersed particles has been explained by the Orowan's mechanism at low temperatures. When the diffusion can easily occur at high temperatures, the matrix can deform plastically at stresses lower than the Orowan stress. It has been proposed by Shewfelt and Brown that a dislocation can overcome dispersed particles by local climb at high temperatures and so the deformation can proceed⁽³⁾. On the other hand, Nardone *et al.*⁽⁴⁾ has explained the threshold stress using a nonmechanical pinning which arises from the attractive interaction between a dislocation and dispersed particles, and it has been shown that the threshold stress

σ_{th} can be lower than the Orowan stress at high temperatures. It is thought in this case that the dislocation lowers the line energy by contacting to the interface between matrix and particles. The line energy is composed of the energies due to the elastic strain field and the dislocation core. In the former case it has been thought that the force acting on a dislocation is different depending on the value of the elastic constant of dispersed particles and matrix, and that the force is repulsive if the elastic constant of particles is greater than that of matrix. However, it has been shown by Srolovitz *et al.*⁽⁵⁾⁽⁶⁾ that if a slipping interface is taken into consideration as the interface between matrix and dispersed particles at high temperatures the attractive force acts on the dislocation even when the particle is harder than the matrix. Moreover, the energy of the dislocation core is expected to decrease if dislocations can be annihilated on the interface between matrix and dispersed particles⁽⁵⁾. Nardone *et al.* has proposed an effect of solute atoms on the interaction between dispersed particles and dislocations. If the solute atoms segregate on the interface they act as pinning to the motion of dislocations. As shown above, some factors can be thought as the origin of the threshold stress for the deformation at high temperatures.

Deformation behaviors of an Al-Mg solid solution alloy and an dispersion hardened Al-Mg-Mn alloy, which was strengthened by Al₆Mn particles, have been discussed by Nakajima *et al.*⁽¹¹⁾⁽¹²⁾. The presence of threshold stress was obviously observed in the Al-Mg-Mn alloy, the value of which was close to the Orowan stress independent of the deformation temperature. It was presumed, in this case, that the threshold stress was

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due to the attractive interaction between dislocations and particles on the basis of the observation of the dislocation structure after deformation.

It has been known that the addition of Zr to Al as a solute element prevents recovery and recrystallization⁽¹³⁾⁽¹⁴⁾. It was shown by the present authors that the addition of Zr to Al-Al₂O₃ dispersion hardened alloys has a great influence upon the creep behavior and raise the creep strength⁽⁸⁾. In this case, the creep behavior could be explained as being controlled by the self-diffusion of Al, taking the temperature dependence of the threshold stress into account.

In the present work, high temperature tensile tests and creep tests were carried out on a dispersion hardened Al-Al₂O₃ alloy with a small addition of Zr, in order to examine the creep behavior and the critical stress in detail.

The material was composed of coarse crystal grains and had an anisotropy of strength due to the texture which arised from recrystallization. The relation between the yield stress and the critical stress of creep deformation was examined to discuss the threshold stress of the dispersion hardened alloy containing a solute element.

II. Experimental Procedure

The material used in the experiment was an Al-Al₂O₃-0.09%Zr alloy which was prepared by the melt sintering method. The chemical analysis of this material is shown in Table 1. Creep specimens and tensile specimens were machined from rolled plates 1 mm in thickness using a spark cutting machine. Three kinds of specimens were prepared which had different tensile axes against the roll-

Fig. 1 Optical micrographs of the specimen after heat treatment. (a) Surface, (b) Long transverse section.

Table 1 Chemical composition (mass%) of the alloy and volume fraction f (%) of Al_2O_3 particles.

Alloy	Zr	O	f
Al- Al_2O_3 -0.09%Zr	0.094	0.33	0.48

Table 2 Grain size after heat treatment.

l/mm	w/mm	t/mm
0.8	0.4	0.05

ing direction. The 0° , 45° and 90° specimens had the tensile axis of 0° , 45° and 90° against the rolling direction, respectively. The specimens were heat-treated at 873 K for 3.6 ks for solution treatment and homogenization after recrystallization at 673 K for 7.2 ks in an air furnace. The grain structures on the surface and the longitudinal section of the plate after the heat treatment are shown in Fig. 1. The crystal grains were coarse and flat in shape, being elongated in the rolling direction and thinned in the thickness direction. The grain sizes measured in the directions of longitude(l), width(w) and thickness(t) are shown in Table 2.

The $\{111\}$ pole figures for the recrystallization textures measured by X-ray diffraction are shown in Fig. 2. The main peaks of the $\{111\}$ pole were observed in four directions, showing the clear recrystallization texture. These peaks had the orientations of about 30° against the normal to the specimen surface and about 75° against the rolling direction.

Creep tests were carried out in tension under a constant stress condition using a lever type creep testing machine in the temperature range from 573 to 873 K. Ten-

sile tests were carried out using an Instron type tensile testing machine at the initial strain rate of about $5 \times 10^{-4} \text{ s}^{-2}$ in the temperature range from room temperature to 773 K.

III. Experimental Results

1. Creep behavior

Relations between the steady state creep rate $\dot{\epsilon}_s$ and the applied stress compensated by Young's modulus are shown in Fig. 3 in the double logarithmic plots. The creep rates of the 0° , 45° and 90° specimens decreased steeply with the decrease in applied stress over the temperature range from 573 to 773 K. These results suggest the existence of threshold stress $(\sigma/E)_c$ for creep deformation. At 873 K fairly high creep rates were observed even at low stresses, not showing the steep decrease in creep rate which was observed at temperatures lower than 773 K. Similar phenomena have been observed in the creep deformation of an Al- Al_2O_3 binary alloy, which may be attributed to the increase in grain boundary sliding and the decrease in threshold stress at high temperatures⁽⁷⁾.

2. Tensile tests

Stress strain curves and their magnified curves around the yield points are shown in Figs. 4(a) to 4(c) and in Fig. 5, respectively. At temperatures from room temperature to 773 K no serration was observed in the stress strain

Fig. 2 Pole figure of the $\{111\}$ plane.

Fig. 3 Stress dependence of steady state creep rate of the specimens with three different specimen axes.

Fig. 4 Stress-strain curves. Specimen axis: (a) 0°, (b) 45°, (c) 90°.

curves. The yield stress, work hardening rate and tensile strength decreased with the increase in temperature. The temperature dependences of the 0.1% proof stress and 0.01% proof stress, which are taken as a measure of the macroscopic and microscopic yield stress respectively, are shown in Fig. 6. The proof stress was generally the highest in the 0° specimen at each temperature and the lowest in the 90° specimen. Though the proof stress of each specimen decreased slowly with the increase in temperature in the range from room temperature to 473 K, they showed a tendency to decrease a little steeply in the temperature range from 473 to 673 K. Around 773 K the proof stress decreased slowly again, and the differences in proof stress among the three types of specimens became very small. These tendencies were the same for the 0.2% proof stress and 0.02% proof stress. In Figs. 6(a) and (b) solid marks denote the critical stresses obtained in the creep tests, as shown later.

Fig. 5 Magnified stress-strain curves of the specimen with the 0° specimen axis.

Fig. 6 Temperature dependence of 0.01% and 0.1% proof stress of the specimens with different specimen axes. (a) 0.01% proof stress, (b) 0.1% proof stress.

IV. Discussion

1. Creep behavior

The creep rate increased with applied stress in Fig. 3 and the relation between the creep rate and stress depended on temperature in each specimen. The relations between the creep rate compensated by the self-diffusion coefficient of Al⁽¹⁶⁾, $\dot{\epsilon}_s/D_{Al}$, and the stress compensated by Young's modulus⁽¹⁷⁾, σ/E , at different temperatures

should be denoted by one curve if the creep deformation is controlled by the self-diffusion of the matrix, as observed in pure metals and Al–Al₂O₃ alloys containing no solute element⁽⁹⁾. The relations between $\log(\dot{\epsilon}_s/D_{Al})$ and $\log(\sigma/E)$ are shown in Fig. 7. The relations cannot be denoted by one curve although the curves at different temperatures approach together more closely compared with those in Fig. 3. It appears that the creep deformation is not controlled by the self-diffusion of Al.

The creep rate did not steeply decrease in each specimen at 873 K. That is, they deformed at a higher rate in the low stress range at 873 K than at the temperatures below 773 K. These results are considered to be due to the increase in the effect of grain boundary sliding in the high temperature, low stress range, as shown previously for an Al–Al₂O₃ alloy⁽⁷⁾. This tendency is also slightly observed in the low stress range at 773 K, as presumed from the comparison of the dash-and-dotted line with the broken line shown in the 45° and 90° specimens. The influence of grain boundary sliding is very little at 773 K in the 0° specimen, in which the least grain boundary sliding is expected from the elongated grain shape.

It can be thought that the creep deformation of the present material is controlled not by the self-diffusion of Al but by the diffusion of solute atoms, *e.g.* Zr, from the result that the relations between $\log(\dot{\epsilon}_s/D_{Al})$ and $\log(\sigma/E)$ are not denoted by one curve in each specimen. The apparent activation energy of creep deformation, Q_c , was calculated from the Arrhenius plots of creep rates at some given stresses in the relations between $\log \dot{\epsilon}_s$ and $\log(\sigma/E)$ (Fig. 3). The activation energy obtained was in the range from 200 to 240 kJ·mol^{−1}, which was relatively close to the activation energy for the diffusion of Zr in Al⁽¹⁵⁾, in the three kinds of specimens, except for the results in the low-stress, high-temperature range. However, as shown in the previous paper, the creep rate of an Al–Al₂O₃–0.02%Zr alloy which contained a smaller amount of Zr was consistent with that of the present material, so it is hardly conceivable that the creep deformation of these alloys is controlled by the diffusion of Zr⁽⁸⁾.

There is another possibility which might explain such a high value of apparent activation energy. That is, it is considered that the apparent activation energy of creep deformation can be larger than that for self-diffusion of Al because the critical stress for creep deformation depends on temperature in each specimen, though the creep deformation is actually controlled by the self-diffusion of Al.

2. Empirical equation for the creep rate and the critical stress

The critical stress is regarded as the threshold stress above which creep deformation can proceed. When the critical stress exists for dislocation creep, the steady state creep rate can be expressed by the *n*-th power law of effective stress ($\sigma - \sigma_c$), where σ is the applied stress and σ_c the critical stress. As reported previously, for some Al–Al₂O₃

Fig. 7 Stress dependence of temperature compensated creep rate, $\dot{\epsilon}_s/D_{Al}$. Specimen axis: (a) 0°, (b) 45°, (c) 90°.

alloys containing no Zr, the steady state creep rate could be expressed by the following empirical equation independent of the volume fraction of the dispersed particles⁽⁹⁾:

$$\frac{\dot{\epsilon}_s}{D_{Al}} = A \cdot \left(\frac{\sigma/E - (\sigma/E)_c}{(\sigma/E)_c} \right)^n, \quad (1)$$

where D_{Al} is the diffusion coefficient of Al, E the Young's modulus of Al and A a constant. The value of stress exponent n , which depends on the grain size, has been estimated to be from 7 to 8 for the coarse grain specimen of the Al-Al₂O₃ alloy. The results obtained in the present experiment were applied to above equation taking the tem-

perature dependence of the critical stress into consideration. Though it was difficult to determine explicitly the critical stresses for creep deformation, $(\sigma/E)_c$, in eq. (1) by experiments, they were estimated for the 0°, 45° and 90° specimens from the data in Fig. 7 in the following way. Since the effect of grain boundary sliding was very small in the 0° specimen, at first the value of $(\sigma/E)_c$ ($=10^{-3.58}$) was determined at 773 K for the 0° specimen so as to make the relation, $\dot{\epsilon}_s/D_{Al} \propto [\sigma/E - (\sigma/E)_c]^n$, hold well for the data at 773 K, where relatively low values of $\dot{\epsilon}_s/D_{Al}$ were obtained, as reported previously⁽⁸⁾. This value is in the range of $(\sigma/E)_c$ estimated by the extrapola-

Table 3 Values of the critical stress at each temperature.

Specimen	Temperature	$\log (\sigma / E)_c$	$(\sigma / E)_c$
0°	773 K	-3.58	2.63×10^{-4}
	673 K	-3.56	2.77×10^{-4}
	573 K	-3.52	3.00×10^{-4}
45°	773 K	-3.61	2.48×10^{-4}
	673 K	-3.59	2.56×10^{-4}
	573 K	-3.55	2.84×10^{-4}
90°	773 K	-3.63	2.33×10^{-4}
	673 K	-3.62	2.42×10^{-4}
	573 K	-3.57	2.68×10^{-4}

tion to the lower stress in the figure, namely, $(\sigma/E)_c = 10^{-3.54} \sim 10^{-3.60}$, and is nearly equal to the value of the binary Al-Al₂O₃ alloy, $(\sigma/E)_c = 10^{-3.56(9)}$. The values of $(\sigma/E)_c$ at 673 K and 573 K were estimated from the value at 773 K and the amount of horizontal movement between the curves at 673 K (or 573 K) and 773 K in Fig. 7. For the 45° and 90° specimens the values of $(\sigma/E)_c$ were estimated at each temperature from the critical stress of the 0° specimen and the average amount of horizontal movements of the curves, because the shape of the curves of these specimens resembled to that of the 0° specimen. The values of $(\sigma/E)_c$ estimated by the above method are shown in Table 3. The relations of eq. (1) using these critical stresses are shown in Fig. 8. In the figure the data at 873 K and the data in the low stress range at 773 K are omitted, because the effect of grain boundary sliding is supposed to be large. The relations for the 0°, 45° and 90° specimens are expressed by the respective straight lines independent of the testing temperature. These straight lines are almost expressed by one relation. Hence it is considered that the creep behavior of the three types of material is expressed by eq. (1) coherently, although the value of the stress exponent and the inclination of the line for each specimen are slightly different. The values of A and n determined for each specimen are shown in Table 4.

3. The temperature dependence of proof stress

The yield stress of dispersion hardened alloys at low temperatures, at which the climbing of dislocations is not expected, has been explained by the Orowan stress. At high temperatures, where the diffusion can easily occur, dislocations can overcome dispersed particles by local climb. The temperature dependence of the Orowan stress is small, which is almost equal to that of the elastic modulus, and it is not dependent on the strain rate. On the other hand, the yield stress depends largely on temperature when the local climb and other processes, which accompany the climb motion of dislocations, operate. The 0.1% and 0.01% proof stresses in Fig. 6 decreased rather steeply with the rise of temperature above 473 K but less depended on temperature around room temperature. Therefore the proof stresses at room temperature can be considered to correspond to the Orowan stress.

The 0° specimen had the highest proof stress and the

90° specimen the lowest one at every temperature from room temperature to 773 K. The anisotropy due to the recrystallization texture can be considered to be the main reason of this result. The present material has a clear recrystallization texture as shown in Fig. 2. The average Schmid factor of each specimen was estimated in the following way. The {111} planes presented by four large peaks in the pole figure of Fig. 2 were regarded as the main active slip planes. The Schmid factors were calculated from the orientational relation between the main slip planes and the tensile axis for all the slip systems which were able to operate in each specimen. The average of these values was used as the Schmid factor of each specimen for convenience. The values of the Schmid factor obtained by this method for the 0°, 45° and 90° specimens were 0.14, 0.18 and 0.26, respectively. The proof stress determined in tensile tests are considered to depend on the Schmid factor. The reciprocal of the Schmid factor, and the 0.1% and the 0.01% proof stresses are shown against the orientation of the tensile axis in Fig. 9. Each value is normalized by the value for the 0° specimen. The reciprocal of the Schmid factor and the proof stress showed qualitatively a similar dependence on the orientation. It is suggested that the applied tensile stress needed for the Orowan's by-passing is affected by the recrystallization texture and is the largest for the 0° specimen and the smallest for the 90° specimen depending on the orientation of the tensile axis.

4. Relation between critical stress and yield stress

The applied stress needed for the Orowan's by-passing, or the proof stress at room temperature is different for the 0°, 45° and 90° specimens being affected by the recrystallization texture. This is supposed to have a large influence on the creep deformation at high temperatures. If the critical stress for creep deformation corresponds to the stress needed for a dislocation to overcome dispersed particles by the climb motion, e.g. the local climb mechanism, the critical stress can be represented by the relation, $(\sigma/E)_c = \alpha \cdot (\sigma_{\text{Orowan}}/E)$, where α is a constant of about 0.4 for the local climb and one order smaller for the extensive climb⁽³⁾. The $(\sigma/E)_c$ is proportional to the Orowan stress in both cases. It is to be expected, therefore, that the proof stress is closely related to the critical stress for creep deformation, since the proof stresses of the 0°, 45° and 90° specimens at room temperature are considered to correspond to the Orowan stress as stated before. The ratio of the critical stress at each temperature to the proof stress at room temperature, which is compensated by Young's modulus at each temperature, $(\sigma/E)_c/(\sigma_{0.1}/E)_{\text{RT}}$ and $(\sigma/E)_c/(\sigma_{0.01}/E)_{\text{RT}}$, is plotted against the orientation of the tensile axis as shown in Fig. 10. The value of the ratio is not dependent on the orientation of the specimen axis and is nearly constant at each temperature. This shows that the critical stress for creep deformation is proportional to the Orowan stress. The ratio ranges from 0.5 to 0.8 at temperatures from 573 to 773 K and shows a tendency to decrease with the increase in temperature. Moreover it is

Fig. 8 Relations between $\dot{\epsilon}_s/D_{Al}$ and $[\sigma/E - (\sigma/E)_c]/(\sigma/E)_c$, where $(\sigma/E)_c$ is the critical stress which depends on the creep temperature. (a) 0° , (b) 45° , (c) 90° .

Table 4 Values of n and A in eq. (1).

Specimen	n	A/m^{-2}
0°	10.0	13.1
45°	9.1	12.8
90°	8.3	12.6

seen in Fig. 6 that the critical stresses for creep are close to the 0.01% proof stresses which correspond to the microscopic yield stresses.

Nakajima *et al.* examined the combined effects of solution hardening and dispersion hardening comparing the deformation behavior of an Al-Mg solid solution alloy and an Al-Mg-Mn dispersion hardened alloy at high temperatures. They revealed that the presence of Al₆Mn

Fig. 9 The dependence of 0.01% and 0.1% proof stress, $\sigma_{0.01}$ and $\sigma_{0.1}$, and the reciprocal of the Schmid factor on the specimen axis.

Fig. 10 The dependence of $(\sigma/E)_c / (\sigma_{0.01}/E)_{RT}$ and $(\sigma/E)_c / (\sigma_{0.1}/E)_{RT}$ on the specimen axis.

dispersion particles contributed only to the increase in the internal stress⁽¹¹⁾. They concluded that the deformation of this alloy was not controlled by the climb motion of dislocations to overcome the dispersed particles because the internal stress due to the presence of the dispersed particles, which was assessed experimentally, did not depend on temperature. Moreover they showed that

when the density of movable dislocations, ρ_m , was so low that the dislocation spacing was larger than the interparticle spacing, the internal stress due to the presence of the dispersed particles and the threshold stress actually observed were close to the Orowan stress. It was presumed from the observation of the dislocation structure after deformation and other results that the threshold stress was not due to the Orowan's mechanism but to a strong attractive interaction between the dislocations and the dispersed particles.

The critical stress σ_c of the present Al-Al₂O₃-Zr alloy was about 0.5~0.8 of the Orowan stress and depended on temperature as shown previously. The temperature dependence of σ_c shows that the interaction between dislocations and dispersed particles changes gradually with the rise of temperature. Though it is presumed that the local climb or other mechanisms which accompany the climbing motion of dislocations operate at these temperatures, the threshold stress needed for the local climb of dislocations has been considered to be independent of temperature. Nardone *et al.*⁽⁴⁾ showed that the threshold stress lower than Orowan stress could be attained by the nonmechanical pinning resulting from the attractive interaction between dislocations and dispersed particles at high temperatures. They suggested that the segregation of impurities on the interface between dispersed particles and matrix could be one of the reasons for the attractive force. If the solute segregation gradually decreases with the rise of temperature the attractive force will decrease. This may explain the decrease in the critical stress with the rise of temperature, which are observed in the present Al-Al₂O₃-Zr alloy[†]. Furthermore it is thought that the mechanism of the local climb is based on the attractive interaction between dislocations and dispersed particles as indicated by Nardone *et al.* Since the temperature dependence of critical stress was not observed in the binary Al-Al₂O₃ alloy which was produced by the melt sintering but did not contain the Zr atom⁽⁹⁾, it is likely that the segregation of impurities on the particle interface, or the formation of the atmosphere of Zr, contributes to the attractive force in the present material. Because the critical stress for creep in the present Al-Al₂O₃-Zr alloy was about 0.5~0.8 of the Orowan stress, it is suggested that the decrease in the line energy of dislocations or the attractive force between dislocations and dispersed particles in this alloy is not so large as in the Al-Mg-Mn alloy examined by Nakajima *et al.* Whether the critical stress for creep is decided by the local climb or by the nonmechanical pinning will depend on the magnitude of attractive force between dislocations and dispersed particles⁽¹⁰⁾. The properties of the interface between Al₂O₃ particles and the Al matrix is expected to have a large influence on the interaction between dispersed particles and dislocations.

[†] It was observed that dislocations were pinned by dispersed particles in the dislocation structure after creep deformation. This suggests that the attractive force exists between the dislocations and the particles⁽⁹⁾.

V. Conclusion

High temperature tensile and creep tests were carried out on the dispersion hardened Al-Al₂O₃-0.09%Zr alloy, which was composed of coarse crystal grains and had an anisotropy of strength due to the recrystallization texture. The proof stress in the tensile test and the critical stress for creep deformation were examined. The main conclusions are as follows.

(1) The creep strength was the highest for the 0° specimen and the lowest for the 90° specimen. The existence of critical stress for creep deformation, σ_c , was recognized at the temperature range from 573 to 773 K where the effect of the grain boundary sliding was small. The critical stress was the largest for the 0° specimen and the smallest for the 90° specimen at each temperature. It decreased with the rise of temperature. The temperature dependence of σ_c was considered to originate in the interaction between dislocations and Zr atoms which segregated on the interface between dispersed particles and matrix.

(2) The stress and temperature dependences of creep rate, controlled by the self-diffusion of Al, were described by an empirical equation for the three orientations of specimens, by taking into account the temperature dependence of the σ_c .

(3) The proof stress in tensile tests was the highest for the 0° specimen and the lowest for the 90° specimen at the temperature range from room temperature to 773 K. This result was discussed in terms of the average value of the Schmid factor of each specimen. The temperature dependence of the proof stress was small at temperatures from room temperature to 473 K for the 0°, 45° and 90° specimens, so the proof stress at room temperature was considered to correspond to the Orowan stress.

(4) A good correlation was observed between the critical stress σ_c and the proof stress at room temperature. The ratio of σ_c to the proof stress was nearly constant independent of the orientation of the specimen axis at each temperature. Though the ratios of the critical stresses to the 0.01% and 0.1% proof stresses approx-

imately ranged from 0.5 to 0.8, the temperature dependence of the ratios was observed, showing a tendency to decrease gradually with the rise of temperature. These results show that the interaction between dislocations and dispersed particles changes with temperature. The change is considered due to the gradual decrease in the attractive force, which is caused by the segregation of solute atoms on the interface between dispersed particles and matrix, with the rise of temperature.

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