High-Temperature Creep and Threshold Stress of Precipitation-Hardened Al-0.7 at%Mn Alloy*

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In order to examine the validity of Srolovitz's attractive interaction theory in dispersion-hardened alloys, high-temperature creep behavior of Al-0.7 at%Mn alloy has been studied in a wide stress range covering both sides of the calculated Orowan stress. It has been found that there is a clear threshold stress for creep deformation at the highest test temperature of 723 K. The stress is approximately equal to the stress estimated from the attractive-interaction hardening, σ_r , which is nearly equal to the Orowan stress in this alloy. TEM observation also suggests that the interaction between dislocations and Al₆Mn particles is of attractive type. Further, it has been shown that the threshold stresses reported so far in dispersion-hardened alloys are well explained by the attractive interaction. Stress dependence of steady-state creep rate is very different between the high and low stress regions, the stress exponent, n , being 14–21 and 4, respectively. The stress dependence of n is commonly observed in dispersion-hardened alloys. Their boundary stress increases with decreasing temperature, and the activation energy for creep deformation, Q_c , in the high stress region is much higher than that for self-diffusion in pure Al, while the value of Q_c in the low stress region is nearly equal to the activation energy for pipe diffusion. These values of n and Q_c are discussed on the basis of available theories.

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I. Introduction

Dispersion hardening is a means to increase the strength not only at room temperature but also at high temperatures by finely dispersing thermally stable second phase particles in the matrix. The upper limit of the yield-stress increase expectable from the hardening is the Orowan stress. If the increase is as large as the Orowan stress, a threshold stress close to the Orowan stress should be clearly observed for creep deformation.

According to previous investigations, however, there are many examples that the strength increase becomes smaller than the Orowan stress at high temperature and the threshold stress is not clearly observed^{(1)–(3)}. It has been considered that at high temperatures where the diffusion rate becomes high, dislocation can climb over the particles under an applied stress much lower than the Orowan stress, and as a result the threshold stress

becomes too low to be clearly observed.

On the other hand, Srolovitz *et al.*⁽⁴⁾⁽⁵⁾ have shown theoretically that when the matrix-particles interface is incoherent, the dislocation-particle interaction becomes attractive at high temperature, because the dislocation stress field at the interface is relaxed due to interface sliding and volume diffusion. If this is the case, the stress required to detach a dislocation of a reduced self energy from the interface is almost the same as the stress for the void hardening, the particles being regarded as voids⁽⁴⁾. When the size and volume fraction of dispersed particles are small, the theoretical void hardening stress is close to the Orowan stress. Then, for an alloy with incoherent particles a threshold stress as high as Orowan stress should be clearly observed and the dislocation climb mechanism will not operate.

The present authors⁽⁶⁾ conducted *in-situ* TEM observations to clarify the dislocation-Be particle interaction in an Al-1.5 vol%Be alloy, and reported that the dislocation-Be particle interaction is attractive. They also confirmed that a clear threshold stress exists and the value agrees with the stress predicted from the void hardening theory⁽⁶⁾. The Al-1.5 vol%Be alloy has the eutectic composition and the solid solubility of Be in Al matrix is negligibly small⁽⁷⁾. As a result, dispersion parameters of particles did not change during long time creep deformation. Therefore, the Al-1.5 vol%Be alloy is ideal as a model material for the study of high-temperature deformation mechanism in dispersion-hardened alloys.

However, precipitation hardened alloy is practically much more important than the eutectic alloy. In such alloys the amount and size distribution of precipitates may change during a creep test depending on time and

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temperature. As a result, the threshold stress may not be observed. Therefore, it is necessary to examine the volume fraction and size distribution of precipitates before and after the creep test in order to confirm the absence or presence of these effects.

In the present study, an Al-Mn alloy, one of the precipitation-hardened alloys, was used. The average interparticle spacing was evaluated from the temperature dependence of the amount of precipitation and the particle size distribution before and after the creep test, and its effect on the creep behavior was examined. Further, the character of the particle-dislocation interaction was identified from TEM observation of the deformation structure.

It was found that the particle-dislocation interaction is of attractive type, a clear threshold stress exists and its value agrees well with the theoretical value of void hardening.

The threshold stress was compared with the reported values of Al-Be⁽⁶⁾ and Al-Mg-Mn⁽⁸⁾⁽⁹⁾ alloys in which the dislocation-particle interaction has been confirmed to be attractive. It was found that the dependence of the threshold stress on the interparticle spacing is the same in the three alloys in agreement with that of theoretical void hardening. However, the activation energy for creep deformation and the stress dependence of steady-state creep rate did not agree with those of the Al-Be and Al-Mg-Mn alloys. These results are described in the following.

II. Experimental Procedure

1. Specimen

An Al-0.7 at%Mn alloy was melted in air from commercial grade Al-10.1 mass%Mn mother alloy and 99.99% high purity aluminum ingot. Ingots of 160 mm in length and 15 mm in diameter were obtained by chill-cast. They were cold-swaged and machined to rod tensile specimens with a gage part of 22 mm in length and 3 mm in diameter. After recrystallization-treatment in air at 673 K for 3.6 ks (1 h), the specimens were heated at 773 K for 10.8 ks (3 h) to stabilize the precipitated particles.

Chemical analysis after the heat treatment showed that Mn content was 0.7 at%. The grains in the specimens were equi-axed and the average grain diameter was about 40 μm , which is almost the same as that in the Al-1.5 vol%Be alloy reported in the previous paper⁽⁶⁾. The dispersed particles were confirmed to be Al₆Mn from the selected area diffraction patterns, as expected from the phase diagram⁽¹⁰⁾.

The crystal structure of Al₆Mn is orthorhombic with lattice constants of $a=6.5 \times 10^{-10}$ m, $b=7.56 \times 10^{-10}$ m and $c=8.87 \times 10^{-10}$ m. The unit cell contains 24 Al atoms and 4 Mn atoms, 28 in total⁽¹⁰⁾. Therefore, the volume fraction of the particles is estimated to be 4.6% when all the Mn atoms are assumed to precipitate as Al₆Mn. However, Mn atoms dissolve into the Al matrix and the solubility depends on temperature. The volume fraction,

Table 1 Volume fraction, f , of Al₆Mn particles in Al-0.7 at%Mn alloy, estimated from the solubility of Mn at various temperatures.

T/K	773	723	673	623	573
Solubility of Mn (at%)	0.21	0.12	0.08	0.06	0.05
f (%)	3.22	3.85	4.08	4.21	4.28

f , of Al₆Mn particles was estimated from the reported solubility curve⁽¹⁰⁾ at the heat-treatment temperature of 773 K and at each test temperature, and shown in Table 1. The value of f changes from 3.22 to 4.28% in a temperature range from 573 to 773 K.

2. Dispersion parameters and Orowan stress

Figure 1 shows a TEM micrograph showing the size distribution of Al₆Mn particles after the heat treatment. The uniformity of the particle distribution is not very high. However, the distribution and the size of particles were not found to change significantly at grain boundaries.

Since no streaks in the selected area diffraction patterns of the particles were observed, the shape of the particles is believed to be lumpish rather than plate-like. For ellipsoid-approximation, each particle volume was determined from the largest and smallest diameters of the particle. Then the particle radius, r , was evaluated from a sphere of the same volume as the ellipsoid. Though the sphere approximation may not be very good for the estimation of the dispersion parameters of ellipsoid particles. However, because the ellipsoid-particles were randomly oriented concerning their axes and the axial ratio were not very large, we believe that the sphere approximation should not be very bad.

Figure 2 shows histograms of the radius distribution of Al₆Mn particles. Figure 2(a) is the histogram before the creep test, and Figs. 2(b) and 2(c) are the histograms after creep tests at 673 K for 39.6 ks (11 h) under 15.7 MPa and at 573 K for 154.8 ks (43 h) under 21.4 MPa, respectively.

The average interparticle spacing (gap between par-

Fig. 1 Transmission electron micrograph showing the size and distribution of Al₆Mn particles in Al-0.7 at%Mn alloy (stabilized at 773 K for 10.8 ks).

Fig. 2 Histogram showing the distribution of radius of Al₆Mn particles at various temperatures. (a) stabilized at 773 K for 10.8 ks, (b) crept at 673 K for 39.6 ks under 15.7 MPa, and (c) crept at 573 K for 154.8 ks under 21.4 MPa.

ticles) on a slip plane, $\bar{\lambda}$ is given by

$$\bar{\lambda} = 1.25\bar{l}_s - 2\bar{r}_s, \quad (1)$$

where $\bar{l}_s$ is the average center-to-center distance of the particles on the plane for square arrangement of particles, and $\bar{r}_s$ is the average radius of particles cut by that plane. The coefficient 1.25 is the converting factor from square to random arrangement⁽¹¹⁾.

In the previous paper⁽⁶⁾, the authors have proposed a new evaluation of $\bar{\lambda}$ taking into account of the size distribution of particles and reported that when the size

distribution is taken into account, the evaluated $\bar{\lambda}$ becomes larger than that when the size distribution is not considered.

According to the new evaluation, $\bar{l}_s$ and $\bar{r}_s$ are given by⁽⁶⁾

$$\bar{l}_s = \sqrt{\frac{2\pi \bar{r}^3}{3f \bar{r}}} = \frac{1}{\sqrt{2\bar{r}N_v}} \quad (2)$$

$$\bar{r}_s = \frac{\pi \bar{r}^2}{4 \bar{r}} \quad (3)$$

where N_v is the number of the particles per unit volume, $\bar{r}$, $\bar{r}^2$ and $\bar{r}^3$ are the averages of r , r^2 and r^3 , respectively, and f is the volume fraction of the particles.

Table 2 shows $\bar{\lambda}$ and the parameters required to evaluate $\bar{\lambda}$, which were calculated from the size distribution shown in Fig. 2. For the evaluation of $\bar{l}_s$, the volume fraction f shown in Table 1 was used. The size distribution was not measured for the specimens crept at 623 K and 723 K, so that the value of $\bar{\lambda}$ in these specimens was estimated by assuming that the size distribution was the same as that stabilized at 773 K. The value of $\bar{\lambda}$ is about 2 times larger than that when the size distribution is not considered.

The modulus compensated Orowan stress, σ_{Or}/E , was evaluated from the following eq. (4), which is a modified form of the Scattergood and Bacon's equation⁽¹²⁾:

$$\frac{\sigma_{Or}}{E} = A \frac{Mb}{4\pi(1+\nu)\bar{\lambda}} \left[\ln \left(\frac{\bar{D}}{r_0} \right) + B \right] \quad (4)^\dagger$$

$A = 1/(1-\nu)$ and $B = 0.6$ for screw dislocation

$A = 1$ and $B = 0.7$ for edge dislocation

Here, b is the magnitude of Burger's vector ($=2.86 \times 10^{-10}$ m), M the Taylor's factor ($=3.06^{(13)}$), ν the Poisson's ratio ($=0.34^{(14)}$). The Young's modulus, E , was evaluated by using the relation of $E=2(1+\nu)G$, where G is the shear modulus and for G were used the values measured by Sutton⁽¹⁵⁾. $\bar{D}$ is the harmonic mean of $2\bar{r}_s$ and $\bar{\lambda}$, r_0 the inner cut-off radius of dislocation core for which b or $3b$ was used. The Orowan stress shown in Table 2 is the geometrical mean of the values for edge and screw dislocations. The range of evaluated Orowan stress comes from the range of assumed inner cut-off radius. As seen in Table 2, the change in dispersion parameters during the creep test, which was worried at

Table 2 Dispersion and related parameters, Orowan stress, σ_{Or} , and the stress required to detach a dislocation from a void pair, σ_v , at various temperatures in Al-0.7 at%Mn alloy.

T/K	$\bar{r}/\mu\text{m}$	$\bar{r}^2/\mu\text{m}^2$	$\bar{r}^3/\mu\text{m}^3$	$\bar{l}_s/\mu\text{m}$	$\bar{r}_s/\mu\text{m}$	$\bar{\lambda}/\mu\text{m}$	σ_{Or}/E	σ_v/E	σ_v/σ_{Or}
773	0.081	0.013	0.0031	1.5	0.12	1.7	$2.8 \times 10^{-4} \sim 2.4 \times 10^{-4}$	$2.4 \times 10^{-4} \sim 2.0 \times 10^{-4}$	0.84
723	—	—	—	1.4	0.12	1.5	$3.1 \times 10^{-4} \sim 2.6 \times 10^{-4}$	$2.6 \times 10^{-4} \sim 2.2 \times 10^{-4}$	0.84
673	0.089	0.012	0.0021	1.1	0.10	1.2	$3.9 \times 10^{-4} \sim 3.3 \times 10^{-4}$	$3.3 \times 10^{-4} \sim 2.8 \times 10^{-4}$	0.84
623	—	—	—	1.3	0.12	1.4	$3.3 \times 10^{-4} \sim 2.8 \times 10^{-4}$	$2.8 \times 10^{-4} \sim 2.3 \times 10^{-4}$	0.84
573	0.11	0.017	0.0035	1.2	0.12	1.3	$3.6 \times 10^{-4} \sim 3.0 \times 10^{-4}$	$3.0 \times 10^{-4} \sim 2.5 \times 10^{-4}$	0.84

[†] In the original Japanese paper, $\bar{l}_s/0.8$ was used instead of $\bar{\lambda}$. The Orowan stresses calculated using $\bar{l}_s/0.8$ and $\bar{\lambda}$ are not much different, because $2\bar{r}_s$ was much smaller than $1.25\bar{l}_s$ in eq. (1).

first, is insignificant and the effect on the Orowan stress can be regarded as within the data scattering.

3. High-temperature creep test

The creep test was conducted in a temperature range from 573 K to 723 K and in a stress range $9.7 \times 10^{-5}E$ to $8.2 \times 10^{-4}E$. The stress range corresponds to a range of 0.32~2.68 in the unit of the Orowan stress, σ_{or}/E , shown in Table 2. The steady-state creep rate at the highest stress was determined from the steady-state deformation obtained by a constant speed tensile test, for which Shimadzu Autograph AG-10TA was used.

The creep test was conducted in a usual way where different specimens were used for different test conditions, and in a special way where a single specimen was used at each temperature and a series of tests were conducted under various stresses. As described later, the results obtained in the two ways agreed well with each other. The details of the creep test were the same as that reported in the previous paper⁽¹⁶⁾.

4. Observation of dislocation-particle interaction

In order to reveal the dislocation-particle interaction, after creep deformation well into the steady-state, specimens were cooled down to room temperature under load by blowing air. Thin plates were cut out from the gage part of the specimens with an acid saw using hydrochloric acid. The plates were electrolytically polished by the twin jet technique to foils for TEM observation.

The electron microscope used was JEOL JEM-2000FX at HVEM Laboratory in Kyushu University. JEOL JEM 1000 at the laboratory was also used for *in-situ* dynamic observation.

III. Experimental Results

1. Creep curve

Examples of the obtained creep curves are shown in Fig. 3. Figure 3(a) shows the stress dependence of creep curves at 723 K, and Fig. 3(b) the temperature dependence of creep curves at $\sigma/E = 3.98 \times 10^{-4}$. Both of them show that the creep deformation goes into the steady state passing through the transient stage of decreasing creep rate. Such behavior is observed also in pure metals. However, the strain up to the steady state is much smaller than that in pure metals, say, smaller than 0.1% under the present experimental conditions. In this connection, the extent of transient creep is about 5% in pure Al in the similar creep rate range⁽¹⁶⁾.

2. Stress and temperature dependence of steady state strain rate

Figure 4 shows the stress dependence of steady-state creep rate, $\dot{\epsilon}_s$, where for the stress is used the modulus-compensated stress, σ/E . The range of Orowan stress given in Table 2, σ_{or}/E , is also shown in Fig. 4. The open marks ($\circ \triangle \square \nabla$) show the datum points obtained by

Fig. 3 Creep curves for Al-0.7 at%Mn alloy. (a) at 723 K under various stresses and (b) under a constant stress of $\sigma/E = 3.98 \times 10^{-4}$ at various temperatures.

Fig. 4 Double logarithmic plot of the stress dependence of steady-state creep rate at various temperatures in Al-0.7 at%Mn alloy. The hatched area represents the range of the Orowan stresses given in Table 2.

Table 3 Stress exponent, n , of steady-state creep rate and activation energy for creep, Q_c , in Al-0.7 at%Mn alloy.

Test temperature	573 K	623 K	673 K	723 K	Q_c (kJ/mol)
High stress region	16	19	21	14	320
Low stress region	4.2	4.1	4.8	4.3	95

using a single specimen at each temperature, where the specimen was at first crept under the lowest stress into the steady state and then the stress level was increased in sequence to measure the steady-state creep rate, $\dot{\epsilon}_s$, at each stress level. The solid marks (●▲■) show the datum points obtained in the usual way where different specimens were used for the different test conditions. Both of them lie on the same curves. This agreement suggests that there should be no effect of the previous deformation on the $\dot{\epsilon}_s - \sigma/E$ relation obtained by the stress-increment method.

As seen from the figure, the creep behavior of the Al-Mn alloy can be divided into two regions, the low stress region where the stress exponent, n , is about 4 and the high stress region where n is 14~21. The values of stress exponent in the two stress regions are shown in Table 3 for each temperature. At the highest temperature of 723 K, the threshold stress, as shown by $\sigma_{th}^{723 K}$ in Fig. 4, is clearly observed and its value is very close to the lower limit of the calculated Orowan stress range. However, at and lower than 673 K, the threshold stress is not clearly observed, in which the transient stress from the high stress region to the low stress region is higher than the threshold stress at 723 K. The transient stress increases as the temperature decreases. As mentioned above, the high-temperature creep behavior of the Al-Mn alloy is different in the two stress regions. However, n in each region does not systematically depend on temperature. Then, at $\sigma/E = 4.00 \times 10^{-4}$ (in the high stress region) and at $\sigma/E = 1.60 \times 10^{-4}$ (in the low stress region), the steady-state creep rate, $\dot{\epsilon}_s$, was plotted against the reciprocal temperature (Arrhenius plot) and shown in Fig. 5.

A good linear relationship holds in each stress region. From the slope, the activation energy is evaluated to be about 320 kJ/mol in the high stress region, which is much larger than 120~145 kJ/mol⁽¹⁷⁾ reported as the activation energy for lattice self diffusion in pure Al and also larger than 120~229 kJ/mol⁽¹⁷⁾ reported as the activation energy for impurity diffusion of Mn in Al. As previously reported⁽⁶⁾, the activation energy for creep in the high stress region of an Al-1.5 vol%Be alloy is almost the same as the activation energy for lattice self diffusion in pure Al, the result of the present alloy being largely different.

It has been known that the activation energy for creep obtained at constant σ is much larger than the activation energy for lattice self diffusion, but the activation energy at constant σ/E agrees with the energy for lattice diffusion⁽¹⁸⁾. Therefore, it is necessary to consider the temperature dependence of E for the correct estimation of the activation energy. In this work, the value of 320 kJ/

Fig. 5 Arrhenius plot of the steady-state creep rate at two stress levels which respectively belong to the high and low stress regions in Al-0.7 at%Mn alloy.

mol mentioned above was obtained at constant σ/E . Therefore, the effect of temperature dependence of E is excluded.

On the other hand, the activation energy for creep in the lower stress region is 95 kJ/mol which is smaller than the activation energy for lattice self diffusion in pure Al, and it is close to 90 kJ/mol⁽¹⁹⁾ reported as the activation energy for pipe diffusion along dislocation line in pure Al. For the grain boundary sliding when the deformation of the grain interior is controlled by the pipe diffusion, the stress exponent of the strain rate has been reported to be 4⁽²⁰⁾⁽²¹⁾, which agrees with the present result. In pure Al, it has been reported that the power law holds in this stress region, and the pipe diffusion controls the deformation⁽²²⁾.

3. Dislocation-particle interaction

Figure 6 shows a high-temperature deformation structure of a specimen crept under 15.7 MPa at 673 K. It is seen that subboundaries adhere to Al₆Mn particles, which means that the interaction between subboundaries and particles is attractive. Such a subboundary-particle interaction was confirmed also by the *in-situ* dynamic observation. These observations suggest that the dislocation-particle interaction should be attractive.

In an Al-Mg-Mn alloy, the dislocation-Al₆Mn particle interaction has been revealed already to be attractive⁽⁸⁾. Although the present alloy differs from the Al-Mg-Mn alloy in respect of no Mg dissolved in the matrix, above results can be considered as an evidence that the dislocation-particle interaction is attractive. In an Al-Be alloy which also showed the attractive interaction⁽⁶⁾, however, no subboundaries were observed. This difference may

Fig. 6 TEM micrograph taken from Al-0.7 at%Mn alloy crept at 673 K under a stress of $\sigma = 15.7$ MPa.

suggest that there should be some essential difference in the high-temperature deformation mechanism between Al-Mn and Al-Be alloys, though the observations are at present too few to be conclusive.

IV. Discussion

1. Activation energy for creep deformation in the high stress region

The activation energy for creep deformation in the high stress region was much larger than that for the lattice self diffusion in pure Al. The reason for this will be discussed.

As shown in Fig. 4, no clear threshold stress was observed in the Al-Mn alloy except the highest test temperature of 723 K. However, if the threshold stress is assumed to decrease as temperature rises, the high activation energy for creep deformation can be explained as a merely apparent energy. But this interpretation is not reasonable as described below.

Yeh *et al.*⁽⁶⁾ showed that when the stress subtracted from σ by σ_{th} was used and modulus compensated, the activation energy for creep at several subtracted stress, $(\sigma - \sigma_{th})/E$, agreed well with that for lattice self diffusion. Similar results have been reported by Lund and Nix⁽¹⁵⁾, Pharr and Nix⁽²³⁾, Horiuchi⁽²⁴⁾ and Lagneborg and Bergman⁽¹⁾.

By assuming that the true activation energy is the same as that for the self diffusion in pure Al, the threshold stresses at 673 K, 623 K and 573 K are adjusted so that this assumption holds. Then the adjusted threshold stress is subtracted from the applied stress. When the steady-state creep rate, $\dot{\epsilon}_s$, is plotted against the subtracted stress, $(\sigma - \sigma_{th})/E$, we obtain Fig. 7, where σ_{th} is the thus assumed threshold stress.

The assumed threshold stresses shown in Fig. 7 are fairly higher than the transient stresses from the high stress region to the low stress region (cf. Fig. 4). If the assumed threshold stresses truly exist, the $\log \dot{\epsilon}_s$ vs $\log (\sigma/E)$ curves should become as shown in Fig. 8. Thus, the

Fig. 7 Steady-state creep rate as a function of the modulus-compensated stress subtracted from σ by σ_{th} which is assumed to depend on temperature in Al-0.7 at%Mn alloy.

Fig. 8 Representation of double logarithmic plot of $\dot{\epsilon}_s$ vs σ/E expected from the assumed temperature dependence of threshold stress in Al-0.7 at%Mn alloy. The arrows indicate the assumed threshold stresses in Fig. 7.

threshold stress should be clearly observed at every test temperature.

The present experiment is believed to be accurate enough to observe such threshold stresses if they truly exist. Actually, however, no clear threshold stress was observed except 723 K. This indicates that the assumption

Fig. 9 Stress dependence of the steady-state creep rate compensated by lattice self-diffusion coefficient, D , of pure Al, in Al-0.7 at%Mn alloy. The curves for pure Al⁽²⁴⁾, Al-1.5 vol%Be alloy⁽⁶⁾ (1 and 2 with different dispersion parameters), and Al-2.77 at%Mg-6.3 vol% Al₆Mn alloy⁽⁸⁾ are shown for comparison.

of the temperature-dependent threshold stress is invalid.

Figure 9 summarizes the relations between the steady-state creep rate compensated by the lattice self diffusion coefficient, $\dot{\epsilon}_s/D$, and the modulus-compensated stress, σ/E , of the present Al-Mn alloy, and Al-Be⁽⁶⁾ and Al-Mg-Mn⁽⁸⁾⁽⁹⁾ alloys reported by the author's group. Here, the lattice self diffusion coefficient, D , in pure Al was used for the Al-Mn and Al-Be alloys; $D_0 = 1.75 \times 10^{-5} \text{ m}^2/\text{s}$, $Q = 135 \text{ kJ/mol}$ in $D = D_0 \exp(-Q/RT)$, where RT is the usual meaning. For the Al-Mg-Mn alloy, the interdiffusion coefficient⁽²⁵⁾ of Mg atoms in Al-Mg alloy was used as D ; $D_0 = 0.34 \times 10^{-4} \text{ m}^2/\text{s}$, $Q = 126 \text{ kJ/mol}$.

In the high stress region, the relation between $\dot{\epsilon}_s/D$ and σ/E is represented by a single curve independent of temperature for both of the Al-Be and Al-Mg-Mn alloys. The relation for the Al-Mn alloy, however, is not represented by a single curve, but clearly depends on temperature. As mentioned above, the activation energy for creep of the Al-Mn alloy is significantly larger than that for the lattice self diffusion in pure Al. Such a high activation energy has also been reported for some other dispersion-hardened alloys⁽²⁶⁾⁽²⁷⁾. However, the reasons proposed for the high activation energy cannot be applied to the present alloy. This reason should be settled in future.

2. Threshold stress in dispersion-hardened alloys

The authors have reported in the previous papers that in Al-Be and Al-Mg-Mn alloys the interaction between dislocations and incoherent particles is attractive and that the observed threshold stress is almost equal to the theoretical value of void hardening proposed by Weeks

et al.⁽²⁸⁾ and Scattergood and Bacon⁽²⁹⁾. This agreement suggests that at the particle-matrix interface the dislocation stress field should be relaxed by diffusion and disappear, as theoretically predicted by Srolovitz *et al.*⁽⁴⁾ In this case, the particles interact with dislocations in the same manner as voids and the stress required to detach a dislocation from the particles is given by an equation with the same form as eq. (4). According to Scattergood and Bacon⁽²⁹⁾, the coefficient A in eq. (4) should be changed to

$$A = \frac{1 + \nu \sin^2 \phi}{1 - \nu} \cos \phi \quad \text{for screw dislocation} \quad (5)$$

$$A = \left(1 - \frac{\nu}{1 - \nu} \sin^2 \phi \right) \cos \phi \quad \text{for edge dislocation}$$

where ϕ is the critical angle for the dislocation to depart from the voids and depends on the Poisson's ratio, ν . Applying the theory⁽²⁹⁾ to the present alloy ($\nu = 0.34$), we obtain as $\phi = 0.82 \text{ rad}$ (47°) for a screw dislocation and 0.33 rad (19°) for an edge dislocation. When the Al₆Mn particles are regarded as the voids and b or $3b$ is used as r_0 in eq. (4), the void hardening stress, σ_v/E , becomes as shown in Table 2. The calculated σ_v/E is not much dependent on temperature and almost the same as the observed threshold stress at 723 K, $\sigma_{th}/E = 2.4 \times 10^{-4}$. It is also to be noted that the void-hardening stress is not much different from the Orowan stress.

As mentioned in introduction, the reported threshold stress in dispersion-hardened alloys is in many cases considerably lower than the Orowan stress. In order to explain the low threshold stress, the local climb⁽²⁾⁽³⁾ and general climb⁽¹⁾ mechanisms have been proposed. However, when the dislocation stress field disappears almost perfectly at the matrix-particle interface, the climb mechanism is hardly expected. Then, instead of the explanation by the climb mechanism, we assume that the reported Orowan stress ($\equiv$ void-hardening stress) should have been overestimated, and reconsider the investigations reporting $\sigma_{th} < \sigma_{Or}$.

As described in the previous paper⁽⁶⁾, the particle size distribution must be taken into account to evaluate the Orowan stress and the void hardening stress, otherwise they will be significantly overestimated⁽⁶⁾. Then, among the reports where the threshold stress was considerably lower than the Orowan stress, we select a paper reported by Schröder and Arzt⁽³⁰⁾ where the particle size distribution was given, and re-estimate the average interparticle spacing, $\bar{\lambda}$, by considering the size distribution. The re-estimated $\bar{\lambda}$ gives a void-hardening stress which agrees well with their observed threshold stress (see Fig. 10). This agreement suggests that the dislocation stress field should disappear almost perfectly at the matrix-particle interface and the particles can be regarded as voids also in their Inconel MA6000 strengthened by Y_2O_3 particles^{††}.

†† In the TEM micrograph shown by Schröder and Arzt⁽³⁰⁾, however, the dislocation contrast is clearly observed at the matrix-particle interface, showing that the stress-field relaxation is far from perfect.

Fig. 10 Relation between the threshold stress, σ_{th} , and attractive-interaction hardening, σ_v , for Al-0.7 at%Mn and other dispersion-hardened alloys including Al-1.5 vol% Be⁽⁶⁾, Al-Mg-Mn⁽⁸⁾⁽⁹⁾, Inconel MA6000⁽¹⁸⁾ and Ni-20Cr-2ThO₂⁽²⁸⁾ alloys.

Figure 10 shows the relationship between the threshold stress, σ_{th}/E , and the void-hardening stress, σ_v/E , given by eq. (5) for all the dispersion-hardened alloys to the author's knowledge, in which the existence of threshold stress has been confirmed and the particle size distribution has also been determined. Among these alloys, for Ni-20Cr-2.5ThO₂ alloy reported by Lund and Nix⁽¹⁸⁾, it is not necessary to re-evaluate the value of $\bar{\lambda}$ or $\bar{l}_s$ because $\bar{l}_s$ was directly evaluated by them from the particle number density per unit area, by using the extraction replica technique. As seen from this figure, the datum points lie around a straight line of $\sigma_{th} = \sigma_v$. This means that the threshold stress for creep is almost the same as the void hardening stress.

For all the cases where the threshold stress for creep does not agree with the Orowan stress, $\sigma_{or} (\cong \sigma_v)$, the dislocation climb mechanism has so far been believed to operate. As described above, however, when the average interparticle spacing is correctly evaluated by taking account of the particle size distribution, the observed threshold stress agrees well with the void hardening stress (not much different from the Orowan stress), so that the threshold stress can be well understood by the Srolovitz's mechanism. Accordingly, it is very probable that the Orowan stress was overestimated in many previous papers. In order to settle the problem whether or not the threshold stress can be truly lower than the Orowan stress, the Orowan stress should be re-examined.

V. Conclusion

In order to clarify the high-temperature creep mechanism of an age-hardened Al-0.7 at%Mn alloy, the temperature dependence of dispersion parameters and the change in dispersion parameters before and after the

creep deformation were investigated together with high-temperature creep tests and TEM observation of the deformation structure. The results obtained are as follows.

(1) The change in dispersion parameters during the creep test is insignificant, lying in the data scattering.

(2) From TEM observations of the deformation structure, it is found that there exist subboundaries and the dislocations composing subboundaries are trapped by the particles on the departure side of the dislocations from the particles. This suggests that the dislocation-particle interaction should be of attractive type.

(3) A threshold stress is clearly observed only at the highest temperature of 723 K. The values of the threshold stress is almost the same as the Orowan stress and the theoretical stress for the attractive interaction proposed by Srolovitz *et al.*, which becomes the same as the void hardening stress when the relaxation of dislocation stress field at the matrix-particle interface occurs perfectly.

(4) The stress dependence of the steady-state creep rate is divided into two regions: low stress region with the stress exponent n of about 4 and high stress region with n of 14–21. The transient stress from the high stress region to the low stress region rises as temperature decreases.

(5) The activation energy obtained from the Arrhenius plot of $\dot{\epsilon}_s$ under a constant stress of σ/E is about 95 kJ/mol in the low stress region and about 320 kJ/mol in the high stress region. The energy in the low stress region is close to the activation energy for the pipe diffusion along dislocation line in pure Al. The energy in the high stress region is much larger than the activation energy for the lattice self diffusion in pure Al, and this high energy can not be explained by the assumption of temperature-dependent threshold stress.

(6) Even in the low stress region, a transient creep is observed, as is observed in pure Al. However, the strain-extent in the transient stage is much smaller in the Al-Mn alloy than in pure Al.

(7) For an Inconel MA 6000 dispersion-hardened alloy, for which the threshold stress has been reported to be lower than the Orowan stress, the average interparticle spacing and the Orowan stress are re-evaluated by considering the particle size distribution. The re-evaluated Orowan stress is as small as the threshold stress, suggesting the Orowan stresses reported so far should have been overestimated.

(8) For the dispersion-hardened alloys in which the threshold stress has been clearly observed and the particle size distribution has been determined or the average interparticle spacing has been measured by the extraction replica technique, the void hardening stresses are evaluated to be almost the same as the threshold stress.

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