Threshold Stress for High-Temperature Creep in Particle Strengthened Al-1.5 vol%Be Alloys*

By Ying-Hwa Yeh**, Hideharu Nakashima***, Hiroaki Kurishita****, Shoji Goto*** and Hideo Yoshinaga***

High temperature creep of dispersion-strengthened Al-1.5 vol%Be alloy has been studied over a stress range $3.1 \times 10^{-5} \sim 5.9 \times 10^{-4}E$ (E : the Young's modulus of Al matrix), which corresponds to $0.40 \sim 5.1\sigma_{or}$ (σ_{or} : the Orowan stress), at temperatures from 473 to 723 K. It is found that there occurs a threshold stress for creep deformation, σ_{th} , and the value of σ_{th} is approximately equal to that of σ_{or} and their ratio, σ_{th}/σ_{or} , is almost independent of the particle size and test temperature. At stresses below σ_{th} , plastic deformation of very slow rates takes place which may be associated with grain-boundary sliding and diffusional creep. *In-situ* HVEM observation reveals that the interaction between dislocation and particles is of attractive type and no dislocation line is visible at the particle/matrix interface, indicating that the stress field of a dislocation is almost completely relaxed at the interface. The observed threshold stress is discussed from the viewpoint of the attractive interaction.

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I. Introduction

Dispersion hardening has been considered to be an effective strengthening method not only at room temperature but also at high temperatures. Many investigations have been carried out on high-temperature deformation of dispersion-hardened alloys⁽¹⁾. According to the previous investigations, a threshold stress[†] for creep deformation exists in some cases⁽²⁾⁻⁽¹¹⁾ and does not exist in other cases⁽¹¹⁾⁻⁽¹⁵⁾. For the practical purpose, absence or presence of the threshold stress is very important, and the reason should be clarified. Fundamentally the essential property of dislocation-particle interaction should be revealed.

For this purpose it is highly desirable to conduct experiment on materials in which the dispersion parameters of particles are well controlled and do not change during deformation, and the mechanical properties of the matrix are well known. As such a material that satisfies the above requirements, Otsuka *et al.*⁽¹⁵⁾ used an Al-0.95 mass%Be alloy close to the eutectic composition⁽¹⁶⁾. In this alloy, the solid solubility of Be in Al is extremely small,

at most 2.5 at ppm and the matrix phase can be regarded as pure aluminum.

The solubility of Al in Be particles is also very limited. The particles with a hcp structure are nearly spherical in shape and the particle distribution in the matrix is highly uniform. Because the alloy has an eutectic composition, the particle distribution can be controlled by heat treatment at high temperature keeping constant volume fraction of particles.

Beryllium has a much higher melting point (1560 K) and a higher shear modulus (about 5.5 times higher) than those of aluminum⁽¹⁷⁾.

Otsuka *et al.*⁽¹⁵⁾⁽¹⁸⁾ used four Al-Be eutectic alloys with different dispersion parameters and studied in detail the high-temperature creep behavior under stresses from $1.34 \times 10^{-4}E$ to $1.12 \times 10^{-3}E$ (E is the Young's modulus of pure aluminum) at temperatures from $0.5T_m$ to $0.69T_m$ (T_m is the absolute melting point of pure aluminum). According to their results, no signs suggesting the existence of threshold stress are observed even in the stress-region lower than the Orowan stress. The present authors⁽¹⁹⁾ also could not observe any signs of threshold stress even under appreciably lower stress levels than the Orowan stress in their preliminary experiment on an Al-Be eutectic alloy. However, the Be particles (hcp) should be incoherent with the matrix (fcc) and the dislocation-particle interaction should become attractive at high temperature. Then it is theoretically expected that there exists a threshold stress as large as the Orowan stress⁽²⁰⁾. In fact, in the Al-2.77 at%Mg-6.3 vol%Al₆Mn dispersion hardened alloy, a threshold stress as expected has been observed⁽¹⁰⁾. Therefore, if in the Al-Be alloy the threshold stress does not exist or is much smaller than the Orowan stress if any, this presents a very important problem on the high-temperature deformation mechanism of

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** Graduate Student, Kyushu University, Kasuga 816, Japan.

*** Department of Materials Science and Technology, Graduate School of Engineering Sciences, Kyushu University 39, Kasuga 816, Japan.

**** Department of Materials Science and Technology, Graduate School of Engineering Sciences, Kyushu University 39, Kasuga 816, Japan. Present address: The Oarai Branch, Institute for Materials Research, Tohoku University, Oarai, Ibaraki 311-13.

† A critical stress near which the strain rate rapidly decreases with the decrease in applied stress and below which the deformation does not practically proceed.

dispersion hardened alloys. Therefore, the deformation behavior of the Al-Be alloy around the Orowan stress should be re-examined.

In order to evaluate the Orowan stress rightly, an exact evaluation of the dispersion parameters (the average interparticle distance and the average particle radius on a slip plane) is necessary. However, in the evaluations conducted so far following problems seem to be contained.

(1) As the average interparticle distance, the distance between particles cut by a plane (slip plane) should be used. However, Otsuka *et al.*⁽¹⁵⁾ seem to have used the spatial average distance. Then, it is probable that the Orowan stress was overestimated.

(2) In general, there is a considerable size distribution of particles. Then, it is necessary to take into account the size distribution in evaluating the average distance on a slip plane. Smaller particles have a lower cutting probability by a plane. If this size dependence of the probability is not taken into account, the Orowan stress is overestimated (see Chapter II-2).

(3) The average particle size is determined by using transmission electron microscopy (TEM), where the observable region is limited to the regions thin enough for the electron beam to transmit. Therefore, relatively large particles, which can not be held in such a thin region, are missed in the observation. This missing leads to an underestimation of the average particle radius and overestimation of number density of particles when the density is estimated from the particle volume fraction. This effect also leads an overestimation of the Orowan stress (see Chapter II-2).

By considering the above remarks, the Orowan stress is evaluated and the Orowan stress is found fairly smaller than those reported by Otsuka *et al.*⁽¹⁵⁾ and evaluated from our preliminary experiment⁽¹⁹⁾. Further, investigating the behavior of steady-state creep around and below the Orowan stress, it is found that a threshold stress clearly exists and that the value agrees approximately with the Orowan stress. However, the *in-situ* observation by high-voltage electron microscopy (HVEM) reveals that the dislocation-particle interaction is not repulsive but attractive as theoretically predicted. In this paper, the experimental results mentioned above will be described in detail and the threshold stress will be discussed from the viewpoint of the attractive interaction between dislocation and particles.

II. Experimental Procedure

1. Specimen preparation

An Al-1.03 mass%(1.5 vol%)Be alloy was melted in air from Al-2.55 mass%Be commercial grade mother alloy and 99.99 mass% pure Al ingots and chill-cast to an ingot 190 mm in length and 15 mm × 160 mm in cross section. The ingot was cold-rolled, cold-swaged and machined into specimens with a gauge part of 22 mm in length and 3 mm in diameter. After recrystallization-treated in air at 673 K for 36 ks, the specimens were aged under

different conditions for 127.8 ks (48 h) or 18 ks (5 h) at 873 K to prepare two kinds of specimens with different particle radii and different interparticle spacings. The specimens aged at 873 K for 172.8 ks will be referred to as alloy A and the specimens aged at 873 K for 18 ks as alloy B. The chemical analysis showed that Be content was 1.03 mass% and the main impurity was 0.025 mass%Fe. The grains in these specimens of A and B were equi-axed and the average grain diameter was about 40 μm in both alloys.

2. Dispersion parameters and Orowan stresses

Figure 1 is a transmission electron micrograph (TEM) showing the size and distribution of Be particles in alloy A. The contrast of Be particles is weak because the permeability of electron beam is much higher in the Be particles than in the matrix. However, it is observed that almost all the particles are spherical and distribute uniformly. Further, no tendency of the different size and distribution of particles was recognized between in the matrix and on the grain boundaries. The dispersed particles were confirmed by the selected area diffraction to be of beryllium single phase.

Figure 2 shows the histogram of radius distribution of particles in alloy B as an example. From the size distribution, the mean particle radius, $\bar{r}$, was determined to be 0.16 μm in alloy B and 0.31 μm in alloy A. As seen in Fig. 2, the particle size distributes in a wide range. Then, the mean interparticle spacing on a slip plane, $\bar{\lambda}$, was evaluated by considering the r -distribution as follows.

In the following, f denotes the volume fraction of dispersed particles, N_v the particle number per unit volume (volume density), N_s the number per unit area on a plane (planer density), r the particle radius for the sphere-approximation, r_s the radius of a particle cut by a slip plane, $\bar{l}_s$ the average interparticle spacing (center-to-center interparticle distance) on the plane, and

$$n(r) dr = N_v F(r) dr,$$

the number of particles with a radius from r to $r+dr$ per unit volume. Here, $n(r)$ is the distribution function of r and $F(r) = n(r)N_v$ the existing probability of those par-

Fig. 1 Transmission electron micrograph showing the size and distribution of Be particles in alloy A (as-prepared).

Fig. 2 Histogram showing the distribution of radius of Be particles in alloy B.

ticles.

Then the planer density of particles with a radius from between r to $r+dr$, dN_s , is given by

$$dN_s = 2rN_v F(r) dr \quad (1)$$

Thus,

$$N_s = \int dN_s = 2\bar{r}N_v \quad (2)$$

Here,

$$\bar{r} = \frac{1}{N_v} \int r n(r) dr = \int r F(r) dr \quad (3)$$

is the average particle radius. If a square arrangement is assumed, the average center-to-center distance is given by

$$\bar{l}_s = \frac{1}{\sqrt{N_s}} = \frac{1}{\sqrt{2\bar{r}N_v}} \quad (4)$$

For a random distribution of particles, the $\bar{l}_s$ in eq. (4) has to be multiplied by 1.25⁽²¹⁾. In the same way, the volume fraction and the average particle radius are related by

$$f = \int \frac{4}{3} \pi r^3 N_v F(r) dr = \frac{4}{3} \pi N_v \bar{r}^3 \quad (5)$$

Here, $\bar{r}^3$ means the average of r^3 .

From eqs. (4) and (5) $\bar{l}_s$ is given by

$$\bar{l}_s = \sqrt{\frac{2\pi \bar{r}^3}{3f}} \quad (6)$$

Denoting the average value of r_s for particles with a radius r by $\bar{r}_s$, the average value of $\bar{r}_s$ for all the particles with various r 's, $\bar{\bar{r}}_s$, given by

$$\bar{\bar{r}}_s = \frac{1}{N_s} \int \bar{r}_s dN_s \quad (7)$$

Using the notations shown in Fig. 3,

$$\begin{aligned} \bar{r}_s &= \frac{1}{r} \int_0^r r_s dx = \frac{1}{r} \int_0^r \sqrt{r^2 - x^2} dx \\ &= r \int_0^{\pi/2} \sin^2 \theta d\theta = \frac{\pi}{4} r. \end{aligned} \quad (8)$$

Fig. 3 Schematic view of the intersection of a slip plane with a circular particle of radius, r .

From eqs. (1), (2), (7) and (8),

$$\begin{aligned} \bar{\bar{r}}_s &= \frac{1}{N_s} \int \frac{\pi}{4} r \cdot 2rN_v F(r) dr = \frac{\pi}{4} \frac{1}{\bar{r}} \int r^2 F(r) dr \\ &= \frac{\pi \bar{r}^2}{4 \bar{r}}. \end{aligned} \quad (9)$$

Here, $\bar{r}^2$ means the average of r^2 . From eqs. (6) and (9), the average interparticle spacing, $\bar{\lambda}$, which should be used to evaluate the Orowan stress, is given by

$$\begin{aligned} \bar{\lambda} &= 1.25 \bar{l}_s - 2\bar{\bar{r}}_s = \frac{1.25}{\sqrt{2\bar{r}N_v}} - \frac{\pi \bar{r}^2}{2 \bar{r}} \\ &= 1.25 \sqrt{\frac{2\pi \bar{r}^3}{3f}} - \frac{\pi \bar{r}^2}{2 \bar{r}}. \end{aligned} \quad (10)$$

Here, 1.25 is the correction factor for the random distribution of particles mentioned above.

In the present investigation, $\bar{r}$, $\bar{r}^2$ and $\bar{r}^3$ were evaluated from the particle size distribution as shown in Fig. 2, and the volume fraction f from the chemical analysis. Then $\bar{\lambda}$ was evaluated by substituting these values for the parameters in eq. (10). Further the interparticle spacing, $\bar{\lambda}'$, when the size distribution is not considered was also evaluated by

$$\bar{\lambda}' = 1.25 \sqrt{\frac{2\pi}{3f}} \bar{r} - \frac{\pi}{2} \bar{r} \quad (11)$$

and compared with $\bar{\lambda}$.

The dispersion and related parameters used in the present study are shown in Table 1. In the present alloys, the value of $\bar{\lambda}$ is seen to be 40% or 100% larger for the case where the size distribution is considered than that for the case where the distribution is not considered.

For the evaluation of Orowan stress, σ_{or} , several theoretical equations have been proposed so far. Among them, the equation proposed by Scattergood and Bacon⁽²²⁾ is considered as the most accurate one. The equation is rewritten as

Table 1 Dispersion and related parameters in alloys A and B in μm .

	$\bar{r}$	$\bar{r}^2$	$\bar{r}^3$	$\bar{l}_s$	$\bar{\bar{r}}_s$	$\bar{\lambda}$	$\bar{\lambda}'$
Alloy A	0.31	0.12	0.06	5.21	0.32	5.9	4.1
Alloy B	0.16	0.04	0.02	3.76	0.20	4.3	2.1

$$\frac{\sigma_{or}}{E} = A \frac{Mb}{4\pi(1+\nu)\bar{\lambda}} \left[\ln \left(\frac{\bar{D}}{r_o} \right) + B \right] \quad (12)$$

$$A = \frac{1}{1-\nu} \text{ and } B = 0.6 \text{ for screw dislocation}$$

$$A = 1 \text{ and } B = 0.7 \text{ for edge dislocation}$$

and used. Here E is the Young's modulus, M the Taylor's factor, b the magnitude of Burger's vector, ν the Poisson's ratio, r_o the inner cut off radius of dislocation and $\bar{D}$ the harmonic mean of $2\bar{r}_s$ and $\bar{\lambda}$. On the rewriting the original equation, the relations of $G = E/2(1+\nu)$ and $\sigma = M\tau$ are used. Here, G is the sheer modulus, σ the tensile stress and τ the resolved shear stress. The Scattergood and Bacon's equation is almost the same as the Ashby's equations⁽²³⁾.

To evaluate σ_{or} , for M , 3.06 was used⁽²⁴⁾, for E , the values measured by Sutton on pure Al single crystal⁽²⁵⁾, for ν , 0.34⁽²⁶⁾ and for r_o , b or $3b$ where b is the magnitude of Burgers vector. The evaluated values are shown in Table 2, in which the mean is the geometrical mean of the values for edge and screw dislocations.

3. Creep test

The creep test was conducted by using a constant load creep machine constructed by ourselves⁽²⁷⁾, in the temperature range from 473 to 723 K ($0.51 \sim 0.78T_m$; where T_m is the absolute melting temperature) and in the stress range from $3.1 \times 10^{-5}E$ to $5.9 \times 10^{-4}E$ ($1.6 \sim 34.9$ MPa). The stress range corresponds to a range of $0.40 \sim 5.1\sigma_{or}$ when used the values of σ_{or} shown in Table 2.

To measure the displacement, LVDT with a high sensitivity of 0.1 mm/V was used, whose signal was recorded with a pen-recorder.

To heat the specimens an electric resistance furnace consisting of three parts was used. The temperature of each part was controlled independently. For the creep test under low stresses, the temperature should be controlled with a high accuracy for a long time. A voltage stabilizer was used to suppress the voltage fluctuation of the electric source. The temperature at the central part of

the furnace was PID-controlled and the temperatures at the upper and lower parts of the furnace were manually controlled with voltage transformers. Thus, the temperature fluctuation during the test and the temperature difference along the gage part of the specimen could be controlled to within ± 1 K.

4. In-situ observation of dislocation-particle interaction

After giving 1% plastic strain at room temperature to introduce dislocations, alloy B was electrolytically polished by a twin jet technique and supplied for *in-situ* TEM observation. The reason for the previous introduction of dislocations is to observe the dislocation-particle interaction. After setting in a heating holder with a two-axes tilt apparatus in the electron microscope, JEOL type 1000 at HVEM Laboratory in Kyushu University, the foil specimen was heated up to 700 K maximum at a rate of about 0.16 K/s (10 K/min) in order to move dislocations by thermal stress and dynamically observe the dislocation-particle interaction. The behavior of dislocation was continuously recorded with VTR. The accelerating voltage was 750 kV, at which no radiation damage was observed up to about 7.2 ks (2 h).

III. Experimental Results

1. Creep curve

Figure 4 shows examples of creep curves at 673 K. The slope of the curves, i.e. the creep rate, is seen abruptly to change at a critical stress level. However, in both regions above and below the critical stress a normal transient creep of pure metal type occurs at first and then a steady state creep is realized. The figure shows only the initial part of creep curves to emphasize the transient stage, though all the tests were conducted until the creep was thought to have proceeded well into the steady state. For example, the curves of ⑥, ⑦ and ⑧ in Fig. 4(a) were continued as long as 5.4×10^3 s (1.5 h), 9.1×10^5 s (10.5d) and 1.0×10^6 s (11.6d), respectively.

2. Stress and temperature dependence of steady state creep rate

Figure 5 shows the double logarithmic plot of steady state creep rate, $\dot{\epsilon}_s$, vs. creep stress, σ , at various temperatures. Here, the stress is normalized by the Young's modulus, E , of pure Al. In the figure, the Orowan stress, σ_{or}^{mean}/E , shown in Table 2 is also indicated by the hatched zone for each alloy. For both alloys A and B, a critical stress (threshold stress) is clearly recognized, below which $\dot{\epsilon}_s$ abruptly decreases, and the stress is close to the Orowan stress independent of temperature. The modulus compensated threshold stress, σ_{th}/E , and its ratio to the Orowan stress are also shown in Table 2. In the higher stress of $\sigma > \sigma_{th}$, a good linear relationship between $\log \dot{\epsilon}_s$ and $\log (\sigma/E)$ holds in both alloys A and B. Then, the following power law holds;

Table 2 Modulus-compensated Orowan stress, threshold stress and the stress required to detach a dislocation from a void pair in alloys A and B.

		Alloy A		Alloy B	
σ_{th}/E		7.2×10^{-5}		9.3×10^{-5}	
		$r_o=b$	$r_o=3b$	$r_o=b$	$r_o=3b$
σ_{or}/E	Screw	1.1×10^{-4}	9.5×10^{-5}	1.4×10^{-4}	1.2×10^{-4}
	Edge	7.3×10^{-5}	6.4×10^{-5}	9.5×10^{-5}	8.2×10^{-5}
	Mean	9.0×10^{-5}	7.8×10^{-5}	1.2×10^{-4}	1.0×10^{-4}
σ_{th}/σ_{or}		0.65 ~ 0.98	0.76 ~ 1.1	0.65 ~ 0.98	0.76 ~ 1.1
σ_v/E	Screw	8.9×10^{-5}	7.6×10^{-5}	1.1×10^{-4}	9.8×10^{-5}
	Edge	6.4×10^{-5}	5.6×10^{-5}	8.4×10^{-5}	7.2×10^{-5}
	Mean	7.6×10^{-5}	6.5×10^{-5}	9.8×10^{-5}	8.4×10^{-5}
σ_{th}/σ_v		0.81 ~ 1.1	0.94 ~ 1.3	0.82 ~ 1.1	0.95 ~ 1.3

Fig. 4 Creep curves at 673 K under various stresses for (a) alloy A and (b) alloy B.

$$\dot{\epsilon}_s \propto \left(\frac{\sigma}{E} \right)^n \quad (13)$$

The stress exponent, n , is also shown in Fig. 5, which tends to decrease as the temperature rises.

Figure 6 shows the Arrhenius plot of the steady state creep rate at a constant stress of $\sigma/E = 2.9 \times 10^{-4}$ in a temperature range from 523 K to 673 K for alloy B, in which a fairly good linear relationship holds between $\log \dot{\epsilon}_s$ and $1/T$. From the slope of this line, the activation energy for the steady state creep deformation, Q_c , is determined to be 142 kJ/mol. As shown in Fig. 5, the temperature dependence of n , except the lowest temperature of 473 K, is so small in alloy B that the value of Q_c does not significantly change in the higher stress region. The value of Q_c lies in a range from 138 to 146 kJ/mol which was obtained by Otsuka *et al.*⁽¹⁵⁾ for the creep deformation of four kinds of Al-Be alloys, and also lies in a range from 120 to 145 kJ/mol which has been reported as the activation energy for lattice self diffusion in pure Al⁽²⁸⁾. Therefore, the lattice self-diffusion coefficient of pure Al, D , was used to compensate the temperature dependence of $\dot{\epsilon}_s$ (where $D_0 = 1.75 \times 10^{-5} \text{ m}^2/\text{s}$, $Q_l = 135 \text{ kJ/mol}$ in $D = D_0 \exp(-Q_l/RT)$).

Compensated strain rate, $\dot{\epsilon}_s/D$, is plotted against σ/E in Fig. 7 by double logarithms. Datum points for alloys A and B lie on their single curves, respectively. The broken line represents the relation reported for pure Al⁽⁷⁾.

Comparing the Al-Be alloys with pure Al, both of the absolute value of $\log(\dot{\epsilon}_s/D)$ and its stress dependence are seen to approach those of pure Al as the stress increases in the higher stress region of $\sigma > \sigma_{th}$, on the other hand,

Fig. 5 Double logarithmic plot of the stress dependence of steady-state creep rate at various temperatures in (a) alloy A and (b) alloy B. The hatched areas represent the range of the mean Orowan stresses given in Table 2.

Fig. 6 Arrhenius plot of the steady-state creep rate at a constant stress of $\sigma/E = 2.9 \times 10^{-4}$ in alloy B.

Fig. 7 Stress dependence of the steady-state creep rate compensated by lattice self-diffusion coefficient, D , of pure Al. The broken line represents the curve for pure Al⁽⁷⁾.

in the lower stress region of $\sigma < \sigma_{th}$, the absolute value for alloys A and B is much smaller than that of pure Al, because the threshold stress does not exist in pure Al; for example, the $\dot{\epsilon}_s/D$ at $\sigma/E = 6 \times 10^{-5}$ is about 4 orders of magnitude lower in the alloys A and B than in pure Al. The stress exponent in the lower stress region is about 2 which is much smaller than the values of 4.9–10.5 in the higher stress region.

Figure 8 shows a double logarithmic plot of $\dot{\epsilon}_s/D$ and

Fig. 8 The diffusion-compensated steady-state creep rate as a function of the modulus-compensated stress subtracted from σ by σ_{th} in alloy A and B.

the modulus-compensated stress subtracted from σ by σ_{th} , $(\sigma - \sigma_{th})/E$. As seen from the figure, the curve for alloys A and B is very close to that for pure Al except the stress region near σ_{th} . This shows that if the subtracted stress is used in the higher stress region of $\sigma > \sigma_{th}$, the stress dependence of $\dot{\epsilon}_s$ in the Al-Be alloys is approximately described by that in pure Al. Similar results have been reported by Lund and Nix⁽³⁾, Pharr and Nix⁽⁴⁾, Horiuchi⁽⁹⁾ and Lagneborg and Bergman⁽²⁹⁾.

3. In-situ observation of dislocation-particle interaction

The dislocations introduced in alloy B by 1% plastic strain at room temperature were winding themselves around Be particles. When heated to about 420 K, mutual annihilation of dislocations of opposite sign began to occur (annihilation at the particle-matrix interface may have occurred, though it was not confirmed), and when the temperature rose above about 585 K, a part of dislocations was observed to be attracted to particles.

Figure 9 shows the HVEM micrographs of alloy B recorded by VTR, which show the successive movements of dislocations interacting with Be particles at 585 K. As indicated by an arrow, the dislocation attracted to a particle in the micrograph of No. 1 or No. 4 bows out to the upper side. No. 3 and No. 5 show a state when the dislocation is just about to detach the particle, and in No. 6 the dislocation has already detached the particle.

As schematically shown in Fig. 10, the dislocation bows out not from the arrival side, but from the departure side. Further, the dislocation branches touching the particle are not in the tangential direction to the particle surface, but almost in the perpendicular direction.

From the disposition of dislocations to become perpendicular to the particle surface, the dislocation-particle interaction is considered to be attractive⁽²⁰⁾⁽³⁰⁾. Furthermore, it should be noted that the dislocation contrast is not observed at the matrix-particle interface.

Fig. 9 *In-situ* HVEM micrographs showing the successive movement of dislocations interacting with Be particles. The arrows indicate particles to be noted.

IV. Discussion

The *in-situ* TEM observation shows that the disloca-

tion-particle interaction in Al-Be alloys is attractive. The attractive interaction characteristic at high temperature is considered to come from the relaxation of dislocation stress field at matrix-particle interface⁽²⁰⁾. Therefore, the

Fig. 10 Schematic illustration of a dislocation bowing out between particles on the departure side of particles.

strength of the interaction is thought to depend on the character of the matrix-particle interface and the easiness of the interface sliding⁽²⁰⁾⁽³⁰⁾.

By assuming that the observed threshold stress comes from the attractive interaction between the dislocation and particles, in other words, the threshold stress is the detaching stress of dislocations lowered in energy at the matrix-particle interface, it will be discussed in the following whether the experimental results can be reasonably explained or not.

In the *in-situ* observation of the dislocation-particle interaction by HVEM, no dislocation contrast such as observed by Schröder and Arzt⁽³¹⁾ is recognized at the matrix-particle interface. This suggests that in the present alloy the relaxation of dislocation stress field at the matrix-particle interface occurs almost perfectly and the core of the dislocation segment falling onto the interface spreads over the whole of the particle, resulting in the appearance as if the dislocation segment disappeared.

Such an interface-relaxation is considered to occur when the interface is incoherent⁽²⁰⁾. In the Al-Be alloys, the matrix is fcc and the Be particles are hcp, and this difference in crystal structure may probably result in an incoherent interface and the perfect interface-relaxation.

For the perfect relaxation, the dislocation-particle interaction (attractive interaction) is thought to be almost the same as the dislocation-void interaction. According to Scattergood and Bacon⁽³²⁾, the stress for dislocations to detach the voids, σ_v , is given by an equation of the same form as eq. (12), provided that the factor A in the equation is changed to

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

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

Here, ϕ is the critical angle for dislocations to detach the voids.

Applying the theory to the present alloys ($\nu=0.34$), ϕ is 0.82 rad (47°) for a screw dislocation and 0.33 rad (19°) for an edge dislocation. Then, A is 1.22 for a screw dislocation and 0.88 for an edge dislocation. The σ_v/E and σ_{th}/σ_v evaluated by using these values of A are shown in Table 2. The values of σ_v are seen almost the same as σ_{th} . Further the values of σ_v for screw and edge dislocations and their geometrical mean are seen close to

the corresponding Orowan stresses, respectively, σ_v/σ_{or} being about 0.8 ~ 0.9. Incidentally, the approximate equation derived by Weeks *et al.*⁽³³⁾ from the interaction energy of a straight screw dislocation with voids gives a lower σ_v by about 20%.

The results that the observed σ_{th} is almost the same as σ_v or σ_{or} well corresponds to the TEM observation, which suggests an almost perfect relaxation of the dislocation stress field at the matrix-particle interface. If the origin of the threshold stress is the attractive interaction between dislocations and particles, the strength of the interaction should depend on the degree of the relaxation of dislocation stress field at the interface. Since the relaxation is controlled by diffusion, the threshold stress should depend on temperature and increase as the temperature rises. In the present experiment, however, no temperature dependence of the modulus compensated threshold stress was recognized (Figs. 5 and 7). This result suggests that the relaxation occurs very rapidly in the test temperature range.

Next, we will discuss about the reason why no threshold stress was observed in the report of Otsuka *et al.*⁽¹⁵⁾. They described that the minimum creep stress was reduced to 0.14 ~ 0.4 σ_{or} (calculated from Table A.1 in their paper⁽¹⁵⁾) for their four kinds of materials with different dispersion parameters. However, they used the spatial average interparticle distance to calculate σ_{or} .

When converted to the average interparticle distance on a plane, $\bar{\lambda}$, the average distance becomes 3 ~ 4 times as large as the values reported by them. Since they did not show the particle size distribution, no correction due to the size distribution can be given to their distances. Considering the size distribution, however, $\bar{\lambda}$ should be still larger, as expected from Table 2. Therefore, the minimum creep stress used by them are probably higher than the threshold stress.

For the lower stress region of $\sigma < \sigma_{th}$, the datum points are too few to discuss the deformation mechanism. However, the weak dependences of creep rate on temperature and stress suggests that the grain boundary diffusional creep (Coble creep⁽³⁴⁾) or grain boundary sliding⁽³⁵⁾ should take part in.

V. Conclusion

In order to clarify the presence or absence of threshold stress in the high-temperature deformation of Al-Be dispersion-hardened alloys, constant load creep tests were conducted in a temperature range from 473 K to 723 K and in a stress range from 1.6 MPa to 34.9 MPa (0.40 ~ 5.1 σ_{or} , where σ_{or} is the Orowan stress) using alloy A (the average radius of particles cut by a slip plane, $\bar{r}_s = 0.32 \mu\text{m}$, the average interparticle distance on a slip plane, $\bar{\lambda} = 5.9 \mu\text{m}$) and alloy B ($\bar{r}_s = 0.2 \mu\text{m}$, $\bar{\lambda} = 4.3 \mu\text{m}$). Further *in-situ* TEM observation was conducted to clarify the dislocation-Be particle interaction. The results obtained are summarized as follows.

(1) In both of the alloys A and B, a clear threshold stress exists and the stress is almost equal to the Orowan

stress. When normalized by the Young's modulus, the threshold stress does not depend on temperature.

(2) From the results of *in-situ* TEM observation, the dislocation-particle interaction is revealed to be attractive. By assuming that the dislocation stress field is completely relaxed at the matrix-particle interface and regarding the particles as voids, the critical stress for the dislocation to detach the voids, σ_v , is estimated to be almost the same as the threshold stress, σ_{th} .

(3) In the higher stress region of $\sigma > \sigma_{th}$, when σ_{th} is subtracted from σ , the stress dependence of $\dot{\epsilon}_s/D$ (D : lattice self-diffusion coefficient in the matrix) almost agrees with that of pure Al.

(4) Even in the lower stress region of $\sigma < \sigma_{th}$, the normal transient creep and steady state creep occur. Although the datum points in the lower stress region are too few to discuss the deformation mechanism quantitatively, the observed weak dependence of creep rate on temperature and stress suggests that the Coble creep and/or the grain boundary sliding should take part in the deformation.

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