Transient Creep Mechanism in Pure Aluminum at High-Temperature*

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In order to clarify the deformation mechanism in the transient creep region, the instantaneous plastic strain and the strain rate have been measured by using pure aluminum from a very early stage of transient creep to the steady state. The creep test has been carried out at temperatures from 623 to 823 K under stresses from 0.81 to 6.7 MPa.

It is found that the instantaneous plastic strain does not depend on temperature; it depends exclusively on stress. At a very early stage of transient creep, the Zener-Hollomon parameter, Z, depends strongly on temperature in the lower temperature region (623–723 K), but it is independent of temperature in the higher temperature region (773–823 K). At a later stage, the Z vs strain curves at lower temperatures converge to a single one, which is the same as that in the higher temperature region, and finally comes into the steady-state.

From the temperature-independence, it is concluded that the main part of the instantaneous plastic strain is produced by the athermal motion of dislocations. From the theoretical analysis based on a dislocation-network model, it is inferred that the creep mechanism in the lower temperature region changes from a process, in which the thermally activated glide of some long dislocation links takes part, to the well-known recovery process at an early stage of transient creep.

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I. Introduction

In pure fcc and bcc metals, it has been shown that the effective stress (thermal stress) contribution to the flow stress is so small that the flow stress is essentially equal to the internal stress (athermal stress)(1)-(4). Accordingly, the change in flow stress is determined by the competition between the increment of internal stress due to plastic deformation and its decrement due to dynamic restoration, and expressed by

$$d\sigma = h d\varepsilon_p - r dt,$$

(1)

as shown by Orowan(5). Here, $d\sigma$, $d\varepsilon_p$ and $dt$ are the changes in flow stress, plastic strain and time, respectively, and $h(=d\sigma/d\varepsilon_p)$ the pure work-hardening rate without the restoring effect and $r(=d\sigma/dt)$ the pure restoring rate without the work-hardening effect. In the case of constant stress creep, from eq. (1) the strain rate $\dot{\varepsilon}_p$ is given by $\dot{\varepsilon}_p = r/h$.

As is well known, the high temperature creep in pure metals proceeds in the order of (1) instantaneous plastic strain due to the sudden application of creep stress, (2) transient stage in which the creep rate monotonously decreases with time, and (3) steady-state stage in which the creep rate is almost constant.

The high temperature deformation mechanism in pure metals has been intensively investigated in the steady-state, in which the strain rate is constant for the creep test and the flow stress is constant for the tensile test. The parameters, $h$ and $r$, have been separately measured by various techniques of abrupt stress change(6)(7), abrupt strain-rate change(7)-(9) and stress relaxation(7)(10)(11). All the results show that the work-hardening rate $h$ clearly depends on temperature and the apparent activation energy of the recovery rate $r$ is much smaller than the activation energy for the lattice self-diffusion, suggesting that the dislocation structure, i.e., dislocation density or
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arrangement or both, depends not only on stress but also on temperature. The temperature-dependent dislocation structure may be developed during the transient creep region. Therefore, it is very important to investigate the transient creep behavior.

Raymond and Dorn(12) used the temperature-corrected time, \( \theta = t \exp\left(-\frac{Q_{S, D}}{RT}\right) \), instead of time \( t \) and showed that the creep curves of pure Al lay on a single curve independent of temperature. Here \( Q_{S, D} \) is the activation energy for the lattice self-diffusion, and \( R \) and \( T \) the usual meanings. They concluded that the deformation mechanism in the transient stage is the same as that in the steady-state. But they could not measure exactly the strain in the very early transient stage.

Oikawa et al.(13) showed that when the applied stress was abruptly changed during steady-state creep, the instantaneous strain involved an appreciable plastic strain. The instantaneous plastic strain depended on temperature for the same stress change(6)(7).

On the other hand, Mitra and McLean(14) showed that the instantaneous plastic strain exclusively depended on stress and it was independent of temperature.

Recently, the present authors(15) showed that under a limited deformation condition the activation energy of the initial creep rate immediately after an instantaneous plastic strain was much smaller than the activation energy of lattice self-diffusion.

In this report, the instantaneous plastic strain and the strain rate are precisely measured from a very early stage of transient creep to the steady state. Further, the deformation mechanism is discussed based on the change of dislocation structure with time. Then the temperature dependence of dislocation structure in the steady-state is inferred.

II. Experimental Method

1. Specimen

An aluminum ingot of 99.99% purity was at first peeled off by about \( 5 \times 10^{-3} \) m in thickness, hot-rolled at about 573 K into a \( 1 \times 10^{-2} \) m thick plate, and then cold-swaged into rods \( 7.2 \times 10^{-3} \) m in diameter. From the rods, tensile specimens shown in Fig. 1 were machined. The specimens were annealed at 843 K for 3.6 ks in air and subjected to the tests. The grains after the annealing were found equiaxed and about \( 3 \times 10^{-4} \) m in average diameter.

2. Tensile creep test

When the applied stress \( \sigma \) is rapidly applied to the specimen, an instantaneous strain \( \Delta \varepsilon_a \) is produced. The instantaneous strain is the sum of the elastic strain \( \varepsilon_a/K \) and plastic one \( \Delta \varepsilon_p \). Therefore, \( \Delta \varepsilon_p \) is expresses by

\[
\Delta \varepsilon_p = \Delta \varepsilon_a - \frac{\sigma_a}{K},
\]

where \( K \) is the apparent Young's modulus of the specimen including the elastic deformation of the machine assembly and given by(16)

\[
\frac{1}{K} = \frac{1}{E} + \frac{S}{lk_M}.
\]

Here, \( E \) is the Young's modulus of the specimen, \( k_M \) the spring constant of the machine, \( S \) the cross section of the specimen gauge part and \( l \) its length. When we use a machine with a known \( k_M \), \( \Delta \varepsilon_p \) is obtained from eqs. (2) and (3).

Tensile creep tests were carried out by using a lever-type testing machine. Figure 2 is the schematic representation of the machine. Figure 3 shows the load dependence of \( k_M \). The spring constant \( k_M \) was measured by a stress-change technique with a dummy specimen of Al-4 at%Mg alloy deformed only elastically(17). The instantaneous plastic strain is separately determined from the total strain by using eq. (2) and the relation of \( k_M \) vs \( F \) shown in Fig. 3.

The plastic flow behavior in an early stage of deformation is strongly affected by specimen
alignment and initial dislocation-structure. In the present work, in order to obtain a good alignment and a constant initial dislocation-structure before the test, the specimens were subjected to a plastic prestrain of about 2% at 823 K and then annealed at 843 K for 3.6 ks in situ under a reduced load 4.5 N (stress 0.3 MPa). The grain size did not change by these prestraining and annealing.

The tests were conducted in a temperature range from 623 to 823 K (0.67 to 0.88 of the melting point of Al) and a stress range from 0.81 to 6.7 MPa (the modulus-compensated stress $\sigma_c/E$ range from $1.7 \times 10^{-5}$ to $1.2 \times 10^{-4}$).

The creep stress was rapidly applied from the reduced stress of 0.3 MPa by the removal of the weight $W_2$ (see Fig. 2) which had been balanced by the weight $W_1$. The time needed for applying the stress was as short as 10–30 ms. The displacement was measured by a LVDT with a sensitivity of $\pm 1 \times 10^5$ V/m (full scale, 0.01 mm), and recorded by a digital wave memory (10 bit) and a Beam Corder of 300 Hz. The wave memory had the high sampling rate of 1000 points per second. When a conventional pen-type recorder is used, the inertia effect of the pen is significant. However, the recorders used in the present work have an enough time-resolution.

For heating the specimen, a resistance furnace consisting of three divided parts was used, the upper and lower parts of which were manually controlled by the rheostats and the middle part was controlled by the PID method. The temperature difference along the gauge part of the specimen was less than $\pm 1$ K, and the temperature fluctuation was less than $\pm 1$ K during an entire creep test.

### III. Experimental Results

#### 1. Instantaneous plastic strain

Figure 4 is an example of recorded displacement vs time curve at a very early stage of transient creep. When a creep stress was rapidly applied to the specimen, two clear bend points appeared on the displacement-time curve. The points are indicated by arrows in Fig. 4. The instantaneous strain $\Delta \varepsilon_p$ was determined from the difference between the displacements at the two bend points.

Figure 5 is the relation between $\Delta \varepsilon_p$ and $\sigma_c/E$. For $E$ is used the relation $E = 2(1 + \nu)G$, here $\nu$ is the Poisson’s ratio and $\nu = 0.345$. For $G$ is used the geometric mean of $(C_{11} - C_{12})/2$ and $C_{44}$, reported by Tallon(10). The datum points of $\Delta \varepsilon_p$ lie on a single curve independent of temperature: the instantaneous plastic strain does...
not depend on temperature but exclusively depends on stress. This means that the main part of the instantaneous plastic strain is produced by the athermal motion of dislocations. Therefore, it is considered that the dislocation structure immediately after applying the creep stress does not depend on temperature and it is determined definitely by stress.

Figure 6 shows the double-logarithmic plots of \( \sigma_c/E \) and \( \Delta \varepsilon_p \). A good linear relationship is observed, and the following equation holds:

\[
\frac{\sigma_c}{K} = 1.3 \times 10^{-3} \Delta \varepsilon_p^{0.46}.
\]  

(4)

If the relation, \( \sigma_c = \alpha G b \sqrt{\rho} \), is assumed between \( \sigma_c \) and dislocation density \( \rho \), \( \rho \) should be proportional to \( \Delta \varepsilon_p \), because the exponent 0.46 is almost equal to 0.5. Here \( \alpha \) is a constant which depends on dislocation arrangement, and \( b \) the magnitude of Burgers vector.

The slope of the line in Fig. 6 is considered to be the value of \( h \) at the very beginning of creep, because the line is independent of temperature, any recovery effect being not included in the slope. This value of \( h \) is referred to as the initial work-hardening rate, \( h_i = (\partial \sigma / \partial \varepsilon_p) \), and from eq. (4) \( h_i \) is expressed by

\[
\frac{h_i}{E} = 2.4 \times 10^{-7} \left( \frac{\sigma_c}{E} \right)^{-1.2}.
\]  

(5)

The stress exponent \((-1.2)\) is equal to the exponent determined by the techniques of stress change\(^{6,7}\), strain-rate change\(^{7-9}\) and stress relaxation\(^{7,10,11}\) in the steady-state. The absolute values of \( h_i \) are, however, about one-tenth of those in the steady-state. This result is in accord with the one obtained by Nakata and Oikawa\(^{19}\), who showed that \( h \) increased rapidly with strain and reached a constant value. This large difference may come from the fact that the dislocation structure is much more inhomogeneous, i.e. cell structure, in the steady state than in the instantaneous strain, and only the dislocations at the places of low internal stress can move in the steady state.

2. Initial creep rate

Figure 7 shows the log-log plots of initial creep rate \( \dot{\varepsilon}_i \) and modulus-compensated stress \( \sigma_c/E \). The rate \( \dot{\varepsilon}_i \) was measured by the slope of smoothed displacement-time curve after the in-
stantaneous strain (see Fig. 4). The time needed for the measurement of the initial strain-rate was about 50 ms.

In the higher stress region, $\log \dot{\varepsilon}_i$ depends linearly on $\log (\sigma_c/E)$ at each temperature, and the relation $\dot{\varepsilon}_i \propto (\sigma_c/E)^n$ holds. The stress exponent $n$ is about 5, which agrees with the reported value determined in the steady-state. The absolute values of $\dot{\varepsilon}_i$, however, are much larger than those of steady-state strain-rate.

On the other hand, in the lower stress region the linearity does not hold, and the curves are convex downward. This difference suggests that the deformation mechanism is different between higher and lower stress regions.

Figure 8 shows the Arrhenius plot of the initial creep rate, $\dot{\varepsilon}_i$, at three normalized stress-levels. In the higher temperature region, $\log \dot{\varepsilon}_i$ depends linearly on inverse temperature at each stress. Approximating the relation by $\dot{\varepsilon}_i \propto \exp (-Q/RT)$, the energy $Q$ is equal to the activation energy for the lattice self-diffusion in Al which is shown as the slope of broken line in Fig. 8. The activation energy of recovery rate should be equal to the activation energy of initial creep rate, because $h_1$ does not depend on temperature. Therefore, it is now clarified that the activation energy of recovery rate equals to that for the lattice self-diffusion.

The reported activation energy of recovery rate in the steady-state is about 2/3 of the activation energy for the lattice self-diffusion. The authors have pointed out that the apparent activation energy of recovery rate is affected by the temperature dependence of dislocation structure. The present result provides an experimental support to the assertion.

In the low temperature region, however, the activation energy of $\dot{\varepsilon}_i$ is much smaller than that for the lattice self-diffusion, even smaller than that for the pipe diffusion along dislocations. The slope corresponding to the activation energy for the pipe diffusion is shown as a chain line in Fig. 8. The energy is about 40 kJ/mol. The experimental, gentle slope suggests that the rate-controlling process in this region is not recovery.

The transient temperature of the activation energy of $\dot{\varepsilon}_i$ increases with the decrease in stress. The mechanism in the lower temperature and stress region will be discussed in the next chapter.

3. Creep-rate change in the transient-creep region

Figure 9 shows the relation between the Zener-Hollomon parameter, $Z[=\dot{\varepsilon} \exp (Q_{S.D.}/RT)]$, and the strain in the transient region at $\sigma_c/E=4.4 \times 10^{-5}$, where $Q_{S.D.}$ is the activation energy for the lattice self-diffusion.
energy for the lattice self-diffusion, 135 kJ/mol. At a very early stage of transient creep, $Z$ depends strongly on temperature in the lower temperature region (623–723 K). However, it is independent of temperature in the higher temperature region (773–823 K). At a later stage, however, $Z$ vs strain curves converge to a single one independent of temperature, and finally comes into the steady state. It is inferred that the creep mechanism in the lower temperature region changes from a process, in which the thermally activated motion of dislocation links takes part, to the well-known recovery process at an early stage of transient creep.

IV. Discussion

1. Deformation mechanism at a very early stage of transient creep

From the present results, it is suggested that the rate controlling process in the lower stress and temperature region is different from the recovery controlled by the lattice self-diffusion or pipe-diffusion. Now we discuss the deformation mechanism from three different viewpoints.

1) The effect of excess vacancies produced by deformation

If excess vacancies produced by plastic deformation are ample, thermal formation of vacancies is not necessary for the dislocation climbing; only the migration energy is necessary. It has been reported that the excess-vacancy concentration $C_{ex}$ is proportional to the plastic strain and is expressed by the relation $C_{ex} = 10^{-4} \Delta \epsilon_p$. In this work, an instantaneous plastic strain ranges from $5 \times 10^{-4}$ to $1 \times 10^{-3}$. The ratio $C_{ex}/C_{th}$ is as small as 1/6–1/200, where $C_{th}$ is the thermal equilibrium vacancy concentration and is expressed by the relation $C_{th} = A \exp (-E_F/kT)$, where the entropy term $A$ is 11 and the formation energy $E_F$ is 72 kJ/mol.

The amount of excess vacancies increases with the increase in plastic strain. Therefore, it is considered that the effect of excess vacancies on the initial strain rate is more significant in the higher stress region. However, only in the lower stress region the activation energy of initial strain rate is much smaller than both of the activation energies for the self-diffusion and for the vacancy-migration (about 60 kJ/mol).

As shown in the above discussion, the rate-determining process is not the climb motion aided by the excess-vacancies.

2) The effect of grain boundary sliding

It has been reported that the grain boundary sliding is in general viscous. Therefore, it cannot be expected that any effect of grain boundary sliding is included in the temperature-independent instantaneous strain.

3) The effect of change in dislocation-structure with time

The change in dislocation structure with time will be discussed based on the dislocation-network growth model.

When the number density of dislocation links with lengths of $l_i$ to $l_i+dl_i$ is denoted by $N(l_i)dl_i$ in unit volume, $N(l_i)$ will be zero for the links of $l_i > l_c = (Gb/\epsilon_p)$ under a resolved shear stress $\tau_s$, because those link dislocations move away rapidly.

Now it is assumed that the internal stress from other dislocations is much smaller than the dislocation line tension. As shown in Fig. 10, the dislocation link AC of a length $l_c < l_c$ will bow out under the creep stress to the arc $ABC$ in equilibrium. If the link further bows out with the aid of thermal activation to the second equilibrium position of $ADC$, it will move away rapidly. The activation energy $U(l_i)$ for the link to glide away is given by

Fig. 10 Schematic representation of thermal activation motion of dislocation bowing out.
where a radius of curvature of the arc $ABC$, $r_p$, is the same as that of the arc $ADC$ and approximately equals to $Gb/2\delta c$. From the relation of $r_p \sin (\theta/2) = l_i/2$, $\theta$ is given by $\theta = 2 \sin^{-1}(l_i/Gb)$ as a function of $\delta c$.

The link-length dependence of the activation energy is shown in Fig. 11 at $\sigma_c/E = 4.4 \times 10^{-5}$. From the relation, $\delta c/G = 2(1 + v)\sigma_c/ME$, where $M (= 3.06)$ is the Taylor factor (25), this value of $\sigma_c/E$ corresponds to $\delta c/G = 3.9 \times 10^{-5}$. The activation energy $U(l_i)$ monotonously increases with the decrease in link length and becomes much larger than the thermal energy $kT$ for $l_i < \delta c - 50b$. The dislocation links in a range from $\delta c$ to $\delta c - 20b$ are thermally activatable.

Accordingly to Friedel (26), the growth rate $v(l_i)$ of a link of length $l_i$ is given by

$$v(l_i) = \frac{G b^2 l_i (\pi - \theta - \sin \theta)}{4}, \quad (6)$$

where $D$ is the coefficient of lattice self-diffusion. It is assumed here that the vacancy diffusion along the dislocation line is very fast compared with bulk diffusion, and the effect of jog density is neglected in eq. (7).

The change in $N(l_i)$ is determined by the competition between the decrement due to thermal-ly activated bowing out process and its increment due to recovery process, as expressed by

$$\frac{dN(l_i)}{dt} = - v_0 \frac{b}{l_i} N(l_i) \exp \left( - \frac{U(l_i)}{kT} \right) \frac{\partial [N(l_i) \epsilon(l_i)]}{\partial l_i}, \quad (8)$$

where $v_0$ is the Debye’s frequency.

If the $l_i$-links are thermally activated and join into other networks, the number of the links shorter than $l_i$ will increase. Assuming that this process occurs at random, however, the probability of the production of long links in the range from $l_i$ to $l_i - 50b$ is so small as $5 \times 10^{-4} - 5 \times 10^{-3}$ at $\tau_c/G = 1.0 \times 10^{-5} - 1 \times 10^{-4}$. Therefore, this increment in thermally activatable links is negligible in eq. (8).

Further, the plastic strain rate determined by the thermally activated motion of dislocation links is given by

$$\dot{\varepsilon}_p = \frac{\bar{A} b v_D}{M} \int_{l_i}^{l_i - 50b} \frac{b}{l_i} N(l) \exp \left( - \frac{U(l_i)}{kT} \right) dl_i, \quad (9)$$

where $\bar{A}$ is the average activation area swept out by a dislocation link.

A critical link length $L_c$ is defined as a $l_i$, the links under which are all too short to be affected by the thermal activation, and is arbitrarily chosen here by a short enough length of $l_i - 50b$. Since the length $L_c$ is approximately equal to $l_i$ at $\tau_c/G = 1 \times 10^{-5} - 1 \times 10^{-4}$, the following equation holds:

$$\tau_c = \frac{Gb}{L_c} \geq \frac{Gb}{l_i}, \quad (10)$$

for the thermally activatable links in the range from $l_i$ to $L_c$. A function $P(l_i)$ is introduced by defining $P(l_i) = N(l_i)/N(L_c)$. We assume that the unit of dislocation link-length is $b$. Supposing the imaginary 3-dimensional dislocation-networks composed of the links of the same length $L_c$, the relation between the mesh unit area and the density of dislocations of length $L_c$ is given by

$$L_c^2 = \frac{\beta}{L_c N(L_c)b / \beta}, \quad (11)$$

Here, $\beta$ is a geometric factor depending on the 3-dimensional network structure. From eqs.
The strain rate is simply expressed by
\[ \dot{\varepsilon} = \frac{\beta \bar{A} \nu_D}{M b^3} \left( \frac{\tau_c}{G} \right)^{\alpha_i} \int_{l_i}^{\infty} P(l_i) \exp \left( -\frac{U(l_i)}{kT} \right) dl_i \quad (12) \]

In the present experiment, the time needed for the creep stress application was in a range from 10 ms to 30 ms, and the time needed for the measurement of the initial strain-rate was about 50 ms after the instantaneous straining.

The function \( P(l_i) \) was iteratively calculated for a time interval of 0.1 ms by using eq. (8). Then, the initial strain rate is numerically calculated by using eq. (12). Here \( D \) is given by \( D=D_0 \exp (-Q/RT) \), where \( D_0=1.75 \times 10^{-5} \) m²/s and \( Q=135 \) kJ/mol.

Figure 12 shows the relation between \( P(l_i) \) and \( l_i \) at \( \sigma_c/E=4.4 \times 10^{-5} \). In the figure, 10 ms corresponds to the time needed for the stress application. The function \( P(l_i) \) in the range from \( l_i \) to \( l_i-20b \) is 0 at every temperature. This shows that these very long links have been thermally activated in 10 ms, because their \( U(l_i)'s \) are much smaller than the thermal energy \( kT \). After 60 ms and 100 ms, no difference in the function \( P(l_i) \) is observed in the higher temperature range, 823–723 K, but a slight difference is observed in the lower temperature range, 673–623 K. This means that the two competitive effects of the thermally activated bowing out and the recovery have already been balanced in the higher temperature range, and the rate-controlling mechanism has changed from the thermally activated motion to the climb motion of dislocations. However, in the lower temperature range, a slight difference in \( P(l_i) \) is observed even after 60 ms and 100 ms.

Figure 13 shows the relation between the calculated strain rate \( \dot{\varepsilon}_p/(\beta \bar{A} \nu_D/Mb^2) \) and the elapsed time \( t \). The strain rate rapidly decreases with time. After 60 ms the strain rate in the higher temperature range of 823–723 K has already reduced to the value in steady-state, but at lower temperatures the strain rate still continues to decrease with time.

Figure 14 shows the Arrhenius plot of experimental and calculated strain rates after 60 ms. The both rates are normalized by the initial strain rate at 823 K and \( \sigma_c/E=4.4 \times 10^{-5} \), in order to remove the unknown parameter, \( \beta \bar{A} \). The stress and temperature dependence of the calculated rates approximately agrees with those of experimental ones.

As shown in the above discussion, the theoretical stress and temperature dependences of the initial strain rate essentially agree with the experimental ones. It is concluded that the creep mechanism changes from the thermally activated bowing out process to the recovery process at a very early stage of transient creep. The thermally activated motion of dislocations...
controls the creep deformation in a very early stage of transient creep. However, most of the creep deformation is determined by the recovery process. The duration time for the bowing out process control becomes shorter as the temperature rises, and eventually shorter than the experimental limit. This is the reason why the activation energy of the initial strain rate agrees with that of lattice self-diffusion in the higher temperature region but becomes much smaller in the lower temperature region.

2. Temperature dependence of dislocation structure

In fcc and bcc pure metals, where the dislocation motion is very fast, the moving path in a crystal will not depend on temperature as far as the dislocation structure is the same. Accordingly, the pure work-hardening rate also will not depend on temperature. However, the experimental values of $h$ during steady-state deformation are dependent on temperature even when the normalized flow stress is the same. This fact shows that the dislocation structure depends on temperature even when the stress during steady-state deformation is the same. Consequently, the apparent activation energy for the recovery process is affected by the temperature dependence of dislocation structure.

The absolute values of $h$ in the steady-state are much larger than those in the transient stage. It is considered that the dislocation structure is much more inhomogeneous in the steady state than in the transient stage, and only the dislocation at the place of low internal stress can move. The mobile dislocation density decreases with inhomogeneity. This may be the difference in $h$ between the two states. In the present work, it is clarified that the activation energy of $r$ agrees with that for the lattice self-diffusion when the dislocation structure is temperature-independent. The development of the dislocation structure during the transient stage is considered to depend on temperature. The development may be affected not only by the lattice diffusion but also the pipe diffusion along dislocations in small mesh sizes and the thermal activation of long dislocation links. These effects may cause the temperature dependence of dislocation structure.

VI. Summary

The instantaneous plastic strain and the strain rate have been measured by using pure aluminum from a very early stage of transient creep to the steady state. The results are as follows.

1) The instantaneous plastic strain does not depend on temperature but exclusively depends on stress. This result shows that the most part of the instantaneous plastic strain is produced by the athermal motion of dislocations. Therefore, it is considered that the dislocation structure immediately after applying the creep stress does not depend on temperature; it depends only on stress.

The initial work-hardening rate, $h_i = (\partial \sigma / \partial \varepsilon_p)$ is expressed by

$$h_i/E = 2.4 \times 10^{-7}(\sigma_c/E)^{-1.2}.$$

The stress exponent ($-1.2$) is equal to those determined by the stress change, strain rate.
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change and stress relaxation tests during the steady-state deformation. But the absolute values of $h_i$ are about one-tenth of those in the steady-state. It is considered that the dislocation structure is much more inhomogeneous in the steady state than in the transient stage.

(2) In the higher stress region the stress exponent of $\dot{\varepsilon}$ is about 5, which is equal to the reported stress exponent of steady-state strain-rate in pure metals. In the higher temperature region the activation energy of the initial strain rate is equal to the activation energy for the lattice self-diffusion. The activation energy of the recovery rate should be equal to the activation energy of initial creep rate, because $h_i$ does not depend on temperature. Therefore, the activation energy of recovery rate should equal that for the lattice self-diffusion.

In the lower stress and temperature region the activation energy of initial strain rate is even smaller than that for pipe-diffusion along dislocations and is about 40 kJ/mol.

(3) At a very early stage of transient creep, the Zener-Hollomon parameter $Z$ depends strongly on temperature in the lower temperature region (623–723 K) but does not depend on temperature in the higher temperature region (773–823 K). At a later stage, the $Z$ vs strain curves converge to a single one, which is the same as that in the higher temperature region, and finally comes into the steady-state. It is inferred that the creep mechanism in the lower temperature region changes from a process, in which the thermally activated glide of dislocation links takes part, to the well-known recovery process at an early stage of transient creep.

(4) On the basis of the network growth model, the strain rate determined by the recovery and thermally activated processes have been calculated. The temperature and stress dependence of calculated strain rate is in good agreement with that of experimental strain rate. This shows that the deformation mechanism is a thermally activated motion of dislocations in the lower stress and temperature region. This mechanism, however, is observed only at a very early stage of transient creep.

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