THE THIRTY-EIGHTH HONDA MEMORIAL LECTURE

High-Temperature Deformation Mechanisms in Metals and Alloys

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I. Introduction

The characteristic feature of high-temperature deformation is that the diffusion effect is significant. Although under low stresses and at low strain rates the deformation may occur solely by diffusion such as Nabarro-Herring creep and Coble creep, the paper presented here will be confined to the deformation induced by dislocation motion.

In pure metals and single phase alloys with weak solution-hardening, the origin of deformation resistance at high temperature is thought to be mainly the dislocation interaction, where the diffusion strongly affect the deformation behavior by bringing about rearrangements of dislocations and decrease in dislocation density. In solution-hardened alloys, the origin of the hardening is thought to be the solute-atmosphere formation around dislocations, where the atmosphere is formed by diffusion. In dispersion-strengthened alloys and precipitation-hardened alloys, the origin of the hardening is the dislocation-dispersoid interaction, where the diffusion plays an important role through altering the climb of dislocations over dispersed particles or the relaxation of stress field of dislocations at the dispersoid-matrix interface. Dispersoid coarsening bringing about the strength-decrease, which is a problem for long-term creep, is also due to diffusion.

As mentioned above, the effect of diffusion to be considered depends on the kind of material. In this paper, the deformation behavior is divided into three main types: pure metal type, solution-hardening type and dispersion hardening type, and the deformation mechanism for each type will be described mainly from our investigations.

II. High-Temperature Deformation Mechanism in Pure Metals

When deformed weakly solution-hardened single phase alloys or pure metals, the stress-strain curves shown in Fig. 1 are obtained. The curves in Fig. 1 were measured by tensile tests for pure vanadium of bcc (a) and pure aluminum of fcc (b), the strain rate being changed by a factor of 10 during deformation. As is well known, the curves are of the work-hardening type and the flow stress tends to saturate as the deformation proceeds. The deformation at the saturated flow stress is often referred to as steady-state deformation, which corresponds to the steady-state creep under a constant applied stress. Both of the strain to the steady-state and the steady-state flow stress decrease as the temperature rises and as the strain-rate decreases. The temperature and strain-rate dependence shows that the high-temperature deformation is a thermally activated process. However, as to the interpretation of the process two theories have been confronted with each other and disputed for a long time. One is the theory of glide-control and assumes the dislocation glide as the thermally activated process, and the other is the theory of recovery-control and assumes that the deformation is controlled by recovery.

In the theory of glide-control the glide motion of screw dislocations dragging jogs on them is assumed as the rate-determining process. Because the resistance of jogs is of short range, we may consider it as shown in Fig. 2(a). The figure is called the force-distance diagram and shows schematically how the resistance varies as the dislocation proceeds. When the range of resistance is narrow such as shown in Fig. 2(a) and the activation energy shown as the hatched area becomes the order of thermal-vibration energy, $kT$, the dislocation can surmount the barrier. When the temperature is risen from $T_1$ to $T_2$, the activation energy increases for the same surmounting
frequency as shown in Fig. 2(b), and the flow stress decreases. For increasing the strain rate (proportional to the surrounding frequency) at the same temperature, the activation energy must be decreased. As a result, the flow stress increases. For the jog-drag of screw dislocations such narrow range resistances stand in a row every atomic distance.

This theory can explain the experimental finding that the activation energy for deformation is equal to that for self-diffusion\(^3\) and further can derive the experimental stress-dependence of the steady-state creep rate by assuming that the vacancy-forming jog density is equal to or higher than the vacancy-absorbing (or interstitial-forming) jog density\(^{60-62}\). Against this assumption, Weertman\(^7\) discussed that the vacancy-forming jog density must be lower. However, the theory has been supported by many researchers.

On the other hand, there must be a resistance due to the long-range interaction of dislocations. As shown in Fig. 2(c), this kind of flow resistance is an athermal one insensitive to the thermal vibration, because the activation energy increases remarkably when the applied stress is lowered only a little than the resistance peak. If the jog-drag theory is the case, there should coexist the thermal resistance due to the jog-drag and the athermal resistance due to the long-range interaction of dislocations. However, at a high temperature where the recovery proceeds rapidly, even when the thermal resistance does not exist, the temperature and strain rate dependences of flow stress as shown in Fig. 1 can occur.

The athermal resistance proportional to the square root of dislocation density increases by plastic deformation and decreases by recovery. Figure 3 is a schematic illustration showing that when the strain rate is increased the resistance peak grows higher (increase in flow stress) because of the higher increasing rate of the resistance and when the temperature is risen the peak lowers (decrease in flow stress) because of the higher recovery rate. This means that the athermal resistance which is almost independent of temperature and strain rate if the dislocation structure and arrangement are the same, does depend on temperature and strain rate through the change in dislocation density and arrangement, and that the temperature and strain-rate dependences are similar to those of thermal resistance. In other words, the experimental facts shown in Fig. 1 do not necessarily mean the existence of thermal stress. When the athermal stress is the most part of flow stress, for the creep deformation proceeding under a constant applied stress the deformation proceeds so as to work harden to compensate the amount of recovery. Accordingly the deformation is a recovery-controlled process.

Designating the resistance due to the long-range interaction of dislocations as internal stress \(\sigma_i\), for the theory of glide-control the stress subtracted by \(\sigma_i\) from the flow stress \(\sigma\), \(\sigma_c = \sigma - \sigma_i\), is the effective stress component for the thermally activated motion of dislocation. For the jog-drag, the increase in jog density by deformation as a result of mutual cutting of dislocations, namely, the increase in effective stress is considered to be the main origin of work hardening. The fact that the steady-state
deformation is easily realized at high temperature may be explained by assuming that the thermal recovery of jog density is accelerated and the density rapidly reaches its steady-state value.

On the other hand, it is well known that the dislocations in deformed pure metals aggregate to form a cell structure and at high temperature the cell wall is adjusted to form a subboundary. It is further known that only a few dislocations move fast during deformation, most of dislocations do not move. These facts suggest that the internal stress component should be large and the effect of recovery should not be neglected. If the effective stress component is negligible, if any, namely, if the recovery theory is valid, all we have to consider is the internal stress. We can easily understand the work hardening as the increase in dislocation density. Generally, the driving force for recovery increases as the internal stress (dislocation density) increases. Therefore, the recovery rate increases as the work hardening proceeds and finally they are balanced to realize the steady-state and the flow stress saturates. As temperature rises, the recovery rate increases. As a result, the strain to the steady-state decreases and the saturated flow stress decreases. Thus, the experimental results as shown in Fig. 1 can be understood by the theory of recovery-control.

The theory of recovery-control was proposed prior to the theory of jog-drag, but appeared to have been taken the place by the jog-drag theory which is directly able to derive the temperature and strain-rate dependences of flow stress. However, many observations showing that the flow stress is proportional to the square root of dislocation density in pure metals and weakly solution-hardened mono-phase alloys are thought to support the recovery-control theory.

The attractive point of the recovery-control theory is the simplicity of the theory, where the parameters to be considered are limited to the two parameters; the easily measurable dislocation density \( \rho \) and the proportional coefficient \( \alpha \) connecting the flow stress with the square root of dislocation density \( (\sigma = \alpha \sqrt{\rho}) \), where \( \alpha \) is the shear modulus and \( b \) is the magnitude of dislocations' Burgers vector). However, as to the conclusive evidence of this theory controversies still continue.

The most convincing way is to determine whether a significant magnitude of the effective stress component exists or not, and various techniques to measure separately the internal and effective stresses have been proposed. Many studies concerning this measurement have been reported, but unfortunately the experimental results did not agree and were divided into two groups: one group claimed that a significant magnitude of effective stress was measured, whereas the other group emphasized not. Then more sophisticated considerations were given, and based on them short-term relaxation, sudden stress-change and sudden strain-rate change tests were conducted. However, for the sake of accuracy controversies still continue as yet. Here, only the result obtained by the sudden strain-rate change technique which the present author believes most reliable will be described.

If a significant effective-stress exists, the effective shear-stress \( \tau_e (= \sigma_e / M \), where \( M \) is the Taylor factor) dependence of the average dislocation velocity may be approximately described by

\[
v = B \tau_e^n = (B / M^n)(\sigma - \sigma_i)^n, \tag{1}\]

where \( B \) is the dislocation mobility. Then the tensile plastic strain rate is given by

\[
\dot{\varepsilon} = (\rho b b / (M^{n+1})(\sigma - \sigma_i)^n, \tag{2}\]

where \( \rho_0 \) is the mobile dislocation density. On the other hand, if the effective stress is negligible, \( \sigma = \sigma_i \), and as a result of competition of work hardening and recovery the change in flow stress may be given by

\[
d\sigma = h \, d\varepsilon - r \, dt, \tag{3}\]

where \( h \) is the pure work-hardening rate without recovery effect and \( r \) is the pure recovery rate without work-hardening effect. Then the strain rate in the case of negligible effective stress may be given by

\[
\dot{\varepsilon} = (\dot{\sigma} + r) / h, \tag{4}\]

Comparing eqs. (2) and (4), it is clear that in the case of significant effective stress \( \dot{\varepsilon} \) does not depend on \( \dot{\sigma} \) and solely depends on \( \sigma \), whereas in the case of negligible effective stress \( \dot{\varepsilon} \) does depend on \( \dot{\sigma} \) and does not explicitly depend on \( \sigma \) (provided that it implicitly depends on \( \sigma \) through \( \sigma_i (= \sigma) \) dependence of \( h \) and \( r \).

Figure 4 is a schematic illustration of stress-strain curves obtained by changing the strain rate during steady-state deformation. Because \( \sigma \) changes by the change in strain rate, it can be verified whether \( \dot{\varepsilon} \) depends on \( \dot{\sigma} \) or not. It should be noted here that the term "strain rate" used for the strain-rate change test is not the true plastic strain rate \( \dot{\varepsilon} \), but the apparent strain rate \( \dot{\varepsilon}_a \) including the elastic deformation of the machine assembly. Denoting the elastic constant of the test system (apparent Young's modulus or combined machine stiffness) by \( K \), the elastic strain rate is given by \( \dot{\varepsilon}_a / K \). Then, the true plastic strain rate can be evaluated from \( \dot{\varepsilon}_a \) and \( \dot{\sigma} \) by

![Fig. 4 Schematic illustration of the strain-rate change and extrapolation method](image-url)
\[ \dot{\varepsilon} = \dot{\varepsilon}_0 - \dot{\varepsilon}_d / K, \quad (5) \]

By inspecting whether the \( \dot{\varepsilon} \) thus evaluated depends on \( \dot{\varepsilon}_d \) or not, it may be determined whether the deformation is recovery control or glide control.

Figure 5 is an example obtained for pure aluminum, where \( \ddot{\varepsilon}_d \) was changed from \( \ddot{\varepsilon}_0 \) to various \( \ddot{\varepsilon}_d \) and the true plastic strain rate \( \dot{\varepsilon}_0 \) after the change is plotted against the fractional change of flow stress \( \Delta \sigma / \sigma_0 \) from the strain-rate change point (which corresponds to the change in dislocation structure after the strain-rate change). \( \sigma_0 \) and \( \dot{\varepsilon}_0 \) are respectively the flow stress and the plastic strain rate in the steady-state before the change. Clearly \( \dot{\varepsilon}_0 \) changes with the change in \( \ddot{\varepsilon}_d \) or \( \dot{\varepsilon}_d \), showing that the deformation is recovery control.

When the dislocation structure is the same, \( r \) and \( h \) at the same temperature should be the same. Then, by eliminating \( \dot{\varepsilon}_0 \) from eqs. (4) and (5), we obtain the following equation.

\[ \dot{\varepsilon} = (K \ddot{\varepsilon}_0 + r) / (K + h), \quad (6) \]

In Fig. 6 the change in plastic strain rate \( \Delta \dot{\varepsilon} \) \( (= \dot{\varepsilon}_0 - \dot{\varepsilon}_d) \) is plotted against \( \ddot{\varepsilon}_d \), where the extrapolated \( \Delta \dot{\varepsilon} \) by \( \Delta \sigma \rightarrow 0 \) well satisfies the predicted proportionality of \( \dot{\varepsilon} \) with \( \ddot{\varepsilon}_d \). As \( |\Delta \sigma| \) increases, the datum points deviate from this linear relationship. This is because \( h \) and \( r \) were changed by the deformation after the strain-rate change.

The above is an example for aluminum. The present author and his coworkers obtained the same conclusion for pure iron and vanadium using other techniques\(^{(16,15)}\). From these investigations the author believes that in general the high-temperature deformation of cubic pure metals is recovery control. If this conclusion is valid, the flow stress is determined by dislocation density \( \rho \). The next problem is to know how \( \rho \) changes with the deformation history. The increase of \( \rho \) by deformation corresponds to the pure work-hardening rate \( h \) and the decrease of \( \rho \) by recovery corresponds to the pure recovery rate \( r \). From this reason \( h \) and \( r \) have been separately measured\(^{(16)}\), and from these values it has been tried to predict \( \rho \) for a given deformation history. However, the theory which is able to describe the variation of \( h \) and \( r \) with the deformation has been yet incomplete, and the prediction of the flow stress by this method has not been succeeded.

III. High-Temperature Deformation Mechanism in Solution-Hardened Alloys

Figure 7 shows stress-strain curves of various solid-solution alloys of aluminum at high temperature\(^{(17)}\). When Mg or Cu is dissolved, the flow stress increases remarkably, and at the same time a work-softening, which is called "high-temperature yield point phenomenon", occurs. By dissolving Zn or Ag aluminum does not harden so much, and the stress-strain curve is of work-hardening type as that of pure aluminum.

The solution hardening of these alloys corresponds well to the change\(^{(18)}\) in lattice constant\(^{(19)}\), suggesting that the solution hardening should arise from the dislocation-solute atom interaction due to the size effect. For such a kind of interaction, solute atoms flow around a moving dislocation in such a manner as shown in Fig. 8, where an edge dislocation with a Burgers vector in the direction of \( x \) is assumed to lie at the origin and move in the \( x \)-direction. Near the origin where the interaction is strong, the flow line is nearly circular like those around a stationary dislocation, because the flow velocity of solute atoms is there much faster than the dislocation velocity. As going away from the dislocation, however, the circle becomes deformed in shape and finally solute atoms are not to flow into the dislocation. As the dislocation velocity increases compared with the flow velocity proportional
to the diffusion coefficient $D$, the region characterized by $y_{\text{max}}$, where solute atoms flow into the dislocation, decreases and eventually it becomes smaller than the interatomic distance $b$, when any solute atmosphere may not be formed around the dislocation.

Figure 9 is a calculated example, showing how the concentration distribution around a moving dislocation changes as the dislocation velocity increases. The concentration in an atomic plane directly beneath the dislocation is shown for simplicity. The solute-atmosphere drag stress $\tau_d$ increases and then decreases passing through a peak due to the change in concentration distribution as the dislocation velocity increases. The velocity corresponding to the peak of $\tau_d$ increases as the temperature rises. The behavior is schematically illustrated in Fig. 10. The broken lines show the behavior in the range where dislocations move too fast for solute atoms to follow them and no atmosphere is formed. This means that the resistance in this region is due to stationary solute atoms and of low-temperature type.

When dislocations move at a velocity $v_1$ shown in Fig. 10, it may be considered that the temperature range of $T<T_2$ is the low-temperature range, and $T_2<T<T_3$ the intermediate-temperature range, and $T_3<T$ the high-temperature range. At $T_3$ in the intermediate-temperature range $\tau_d$ decreases as the dislocation velocity increases.
plastic deformation will not occur, only the strain rate will change. Figure 11 gives an evidence of this. The instantaneous strain-change $\Delta \varepsilon$ (apparent strain) due to the stress change $\Delta \sigma$ is proportional to $\Delta \sigma$, and that the slope (proportional constant) equals the elastic constant $K$. This means that the most part of $\Delta \varepsilon$ is the elastic strain and almost no instantaneous plastic strain occurs.

Figure 12 is an experimental result showing that the stress exponent is indeed 1: the change of plastic strain rate $\dot{\varepsilon}$ is proportional to $\Delta \sigma$.

According to $m = 1$, $\sigma_e$ can be determined by

$$\sigma_e = \dot{\varepsilon} \cdot \Delta \sigma / \Delta \dot{\varepsilon},$$

from the slope in Fig. 12. The dislocation density calculated by using $\sigma_e$ thus determined and the theoretical value of mobility $B$ roughly agrees with the density directly measured by the transmission electron microscopy. Thus, the effective stress is considered to be the solute-atmosphere drag stress.

The ratio of the drag stress to the flow stress has been known to increase as the flow stress increases (strain rate increases) at the same temperature and as the tempera-

resulting in a plastic instability. Serrated stress-strain curves (Portevin-LeChatelier effect) observed in the intermediate-temperature range for solution-hardened alloys are thought to be caused by this plastic instability. When temperature is high enough, $\tau_e$ becomes proportional to $\varepsilon$. This means that the stress exponent $m$ in eq. (1) becomes 1.

When the high-temperature deformation is controlled by the dislocation motion dragging the solute atmosphere, on a sudden change of applied stress, as far as the changed stress does not exceed the break-away stress of dislocations from their atmosphere, an instantaneous

Fig. 11 An experimental result obtained by applying the stress-change method to an Al-5.7 at%Mg alloy during steady-state creep at 623 K under a creep stress of 13.5 MPa.

Fig. 12 Proportional dependence of $\Delta \dot{\varepsilon}$ on $\Delta \sigma$ measured in an Al-5.7 at%Mg alloy. The stress-change test was conducted during steady-state creep at 673 K under a creep stress of 4.5 MPa.

Fig. 13 A dislocation structure observed when an Al-5 at%Mg alloy was crept into the steady-state at 573 K under a creep stress of 50 MPa. (a) 002 reflection, (b) 020 reflection.
ture rises at the same flow stress, and it can attain to as large as 60% in an Al-5.7 at%Mg alloy\(^{(22)}\). The internal stress component is proportional to the square root of dislocation density, suggesting that the origin of the internal stress should be the long range dislocation-interaction\(^{(23)}\).

Figure 13 shows a dislocation structure after a high-temperature deformation. Reflecting the high drag stress, the dislocation distribution is highly uniform in contrast with the heterogeneous distribution in pure metals. In substitutional alloys solute atoms must interact with an edge dislocation much stronger than a screw dislocation. However, in Fig. 13 straight edge dislocation lines are scarce and many bowing out dislocations are seen. From such observations a theory called dislocation loop model has been proposed\(^{(23)}\). However, as shown in Fig. 14, these are not dislocation loops but dislocation arcs\(^{(24)}\).

In a photograph Fig. 14(c) printed by superposing Fig. 14(a) and (b), all the \((a/2)\langle 110\rangle\) type dislocations are to be contrasted. Directions of Burgers vectors determined by the no contrast conditions are labelled to each dislocation segment in Fig. 14(c), which shows that even the dislocations looking like loops under some reflecting conditions are actually dislocation arcs with different Burgers vectors linked with one another. Figure 14(d) schematically illustrates an attractive junction formed by dislocation reaction. The dislocation arcs observed are bow-out dislocations between two attractive junctions. The steps on dislocations indicated by arrows in Fig. 13 may correspond to such junctions. Such steps are frequently observed also in other places than those indicated by arrows.

The estimated stress for dislocations to break away from the junctions is nearly equal to the measured internal stress, though a little lower\(^{(24)}\). Accordingly, the origin of the internal stress is considered to be such a dislocation interaction, and the reason for many dislocation arcs may be this interaction.

For a constant strain rate, the average dislocation-velocity \(v\) decreases inversely proportionally as \(\rho\) increases, and as a result the effective stress \(\sigma_e\) for the solute-atmosphere dragging, which is proportional to \(v\), decreases in parallel. On the other hand, the internal stress \(\sigma_i\) due to the dislocation interaction increases proportionally to \(\sqrt{\rho}\). When the decrement in \(\sigma_e\) surpasses the increment in \(\sigma_i\), a work softening will occur. Figure 15 is a schematic illustration of this, showing how \(\sigma_e\), \(\sigma_i\) and \(\sigma\) vary as the dislocation density increases. \(\sigma_e\) increases proportionally to \(\varepsilon/B\), where the dislocation mobility \(B\) increases proportionally to the diffusion coefficient of solute atoms and decreases as the solute concentration increases and as the size effect is stronger\(^{(23)}\). Denoting the

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Fig. 14  Attractive interaction of dislocations on different slip planes\(^{(24)}\). (a) 020 reflection, (b) 002 reflection, (c) superposed (a) and (b), and (d) illustration of an attractive junction.
initial dislocation density by $\rho_0$ and the saturated dislocation density in the steady-state by $\rho_s$, the flow stress $\sigma$ for a large $\dot{\varepsilon}/B$ decreases monotonously with the increase in $\rho$ to the steady-state flow stress. The high-temperature yield point phenomenon as shown in Fig. 7 can be understood in this way. On the other hand, for a small $\dot{\varepsilon}/B$, as shown by broken lines in Fig. 15, the stress-strain curve will be of work-hardening type similar to that of pure metals. The work-softening occurs when the effective stress is larger than 1/3 of flow stress (10).

If the mechanisms by which the flow stress originates are valid, the flow stress can be theoretically predicted for a given dislocation density under any deformation conditions where the solute-atmosphere drag mechanism operates. The problem is to predict the dislocation density for an arbitrarily given deformation path. Some prediction methods have been developed, though they are still semi-empirical (26) (27). A successful example is shown in Fig. 16.

IV. High-Temperature Deformation Mechanism in Dispersion-Strengthened Alloys

Figure 17 is an example showing the effect of dispersed particles on the stress-strain curve at high temperature, where pure niobium and an alloy containing 2.4 vol% ZrO$_2$ particles (the average particle diameter was 0.17 $\mu$m) were used. As seen in this example the flow stress in a dispersion-strengthened material tends to saturate at high temperature earlier than that in a pure metal.

The most important for engineering application of dispersion-strengthened materials may be that there is a threshold stress for high-temperature creep. Figure 18 shows an experimental result obtained by the creep test using two Al-Be alloys containing 1.5 vol% particles, where the steady-state creep rate divided by the self-diffusion coefficient $\dot{\varepsilon}_s/D$ is double-logarithmically plotted against the Young’s modulus-compensated creep stress $\sigma/E$. Using parameters normalized in this way, the datum points at any temperature can be laid on a single master curve for each alloy. In either of the two alloys the creep rate steeply drops at a certain stress. This critical stress is the threshold stress. The average particle radius was 0.31 $\mu$m in Alloy A and 0.16 $\mu$m in Alloy B. Compared with Alloy A, the threshold stress is higher in Alloy B with a smaller particle size and thus a narrower interparticle spacing on a slip plane.

Figure 19 illustrates the models proposed so far as the threshold stress originating mechanism. Figure 19(a) is a model proposed by Orowan (30), where a dislocation loop is left around a particle after a dislocation passed through the particles. The threshold stress originated by this mechanism is referred to as the Orowan stress. However, at high temperature dislocations can climb over the particles as shown in Fig. 19(b) (31) (32). For this climb mechanism, that the dislocation must be elongated on the climbing over yields a resistance for the disloca-
tion to pass the particles. However, the resistance can be lowered as low as 4% of the Orowan stress. This means that there is practically no threshold stress. On the other hand, Srolovitz et al. proposed that in the case of particles incoherent with the matrix, the tractions at the surface of particles due to a dislocation should be relaxed, the shear component of the traction by interface sliding and the normal component by bulk diffusion. If a full relaxation occurs, the traction becomes zero. This means that the particles are felt by dislocations like voids and the dislocations are attracted by the particles due to the image force. In this case, the stress needed to detach the dislocation from the particles gives the threshold stress as shown in Fig. 19(c). The threshold stress equals the void-hardening stress and is as large as 85% of the Orowan stress.

Figure 20 shows threshold stresses measured in alloys strengthened with incoherent particles plotted against the theoretical void-hardening stresses (both values are normalized by the Young’s modulus $E$), showing that they agree with each other. Figure 21 is an electron transmission micrograph showing that dislocations are indeed attracted by particles. The observation as shown in Fig. 17 that the flow stress in a dispersion-strengthened alloy saturates at high temperature after a little deformation may be due to the annihilation of dislocations at the particle surface.

Fig. 18 Stress-dependence of the steady-state creep rate for two Al-1.5 vol%Be alloys with different dispersed-particle sizes. $\bar{\sigma}_o$ is the geometrical mean of Orowan stresses for screw and edge dislocations.

Fig. 19 Schematic illustration of the threshold stress originating mechanisms proposed so far. (a) Orowan mechanism, (b) climb-over mechanism, and (c) Srolovitz mechanism.

Fig. 20 Comparison of experimental threshold stress $\sigma_{th}$ and theoretical void hardening stress $\sigma_{v}$ in various alloys strengthened by incoherent particles.
There are reports showing that the relaxation at the particle surface is incomplete. If this is the case, however, the threshold stress must be different from the void-hardening stress depending on the degree of incompleteness. According to our studies, however, the contrast of dislocations almost disappears on the particle surface, namely, the relaxation is almost complete, no example of a highly incomplete relaxation has been observed. Further there has been proposed a theory which assumes thermally activated break-away of dislocations from the particles. If this is the case, the threshold stress should decrease as the temperature rises and the reduction of creep rate around the threshold stress should not be so steep as that shown in Fig. 18, but more gentle. However, such a gentle reduction seems to have been observed only in materials with an unstable internal structure such as mechanically alloyed materials.

On the high-temperature deformation behavior of dispersion-strengthened alloys many papers have been reported so far. Papers reporting that the threshold stress was as low as about 30% of Orowan stress or too low to be observed are many, and papers reporting a threshold stress as high as the Orowan stress are rather few. Both of the Orowan stress and the void-hardening stress increase inversely proportionally to the average spacing between the particles on the slip plane. Therefore the evaluated value depends on the spacing. According to our inspection, in the papers reporting a much lower threshold stress than the Orowan stress, there seems to be a problem for the estimation of the average particle spacing or the Orowan stress. Further, when the dispersoid grows during creep, or when a high solution-hardening

coeexists with the dispersion-strengthening, no clear threshold stress can be observed. Figure 20 does not contain the data which are considered to involve such a problem.

The above description is the case of incoherent dispersoid. On the other hand, in the case of coherent dispersoid such as Al₃Li or Al₃Zr in aluminum, the Srolovitz mechanism is hardly considered to operate, and the climb-over mechanism is supposed to operate. This suggests that the threshold stress should be too low to be observed. However, our recent investigation showed that the threshold stress was as high as the Orowan stress. If the reason is that some unevenness in the particle shape hinders the climbing-over, is a problem to be settled hereafter.

V. Closing Remarks

It has been described that the high-temperature deformation mechanisms in metals and alloys are considered to be the recovery-control in pure metals, the solute-atmosphere drag in solution-hardened alloys and the Srolovitz mechanism in dispersion-strengthened alloys with incoherent dispersoid. For predicting the deformation-history dependence of the flow stress based on these mechanisms, the history dependence of dislocation structure has to be predicted. The prediction is most difficult in pure metals seemingly most simple, and it seems to be rather simple in solution-hardened alloys. In dispersion-hardened alloys also, when added solution-hardening, it seems to be more simple. However, though it appears simple, the prediction method is yet at a semi-empirical stage, and has not been extended yet to the deformation conditions for the dislocations to break-away from the solute atmosphere. To develop the prediction method is a problem to be solved hereafter. However, the basic knowledge for the prediction seems to have been almost established. While the cooperation of many researchers will be needed hereafter, the author heartily appreciate the assistance rendered by his coworkers.

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