Effect of Particle Shape on the High Temperature Yield Strength of Dispersion-Hardened Nickel Base Alloys*

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In order to investigate the shape effect of particles on the high-temperature strength of dispersion-hardened alloys, two kinds of nickel alloys with SiO$_2$ particles of complex and spherical shapes were made by internal oxidation and a heat-treatment after the internal oxidation, respectively. Their strengths were measured by means of tensile test at high temperatures.

The modulus-compensated yield stress increment due to the complex-shaped particles depended neither on the temperature nor on the strain rate, and agreed with the Orowan stress calculated from the observed dispersion. This fact suggests that the effect of dislocation climb over the particles is effectively suppressed by making the particle shape complex. On the other hand, the yield stress increment due to the spherical particles agreed with the Orowan stress at room temperature but decreased with increasing temperature and decreasing strain rate at 1173 K. This fact suggests that the effect of dislocation climb is significant on the spherical particles. Because of the higher planar number-density of particles, the Orowan stress of the alloy with complex-shaped particles was always higher than that with spherical ones. From the result mentioned above, the complex-shaped particle is found to be effective for dispersion hardening at high temperatures.

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I. Introduction

The yield stress of dispersion-hardened alloys has been well known to be equal to the Orowan stress at lower temperatures below about $T_m/2$, where $T_m$ is the absolute melting point of the matrix metal$^{[1][2]}$. In this case, the dislocations bow out between the dispersed particles on their glide plane under the applied stress, as shown in Fig. 1(a), and then pass through these obstacles, leaving dislocation loops encircling the particles. The value of yield stress evaluated from this dislocation motion is given by about 0.81 $Gb/L$, where $G$ is the shear modulus, $b$ the magnitude of Burgers vector of dislocations and $L$ the mean planar interparticle spacing$^{[3]}$.

At higher temperatures, however, the bulk-diffusion rate will be high and dislocations have been considered to climb over the particles below the Orowan stress, as shown in Fig. 1(b). For the climbing, two mechanisms have been proposed so far. One is called local climb$^{[4]}$, in which the dislocation segment between the particles remains in the glide plane and the remainder profiles the particle surface as it climbs. The other is called general climb$^{[5][6]}$, in which all of the dislocation line climbs out of the glide plane. The lower limit of the yield stress has been evaluated to be

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about 0.32 $Gb/L^{(4)}$ for the local climb and about 0.04 $Gb/L^{(6)}$ for the general climb.

Therefore, the yield stress of dispersion-hardened alloys (DH alloys) should decrease at high temperatures. Many experiments on the yield strength of DH alloys have been conducted at high temperatures so far$^{(7)(8)}$, and their yield stresses have been shown to decrease with the increase in temperature even if the temperature dependence of the shear modulus is compensated. For example, the yield stress of Cu–Al$_2$O$_3$, Cu–BeO and Cu–SiO$_2$ alloys at 1173 K is about half of the stress at room temperature$^{(8)}$.

If the volume fraction of dispersed particles is the same, the decrease of yield strength is expected to be most significant in the case of spherical particles, because the mean climb-distance becomes the minimum for the shape. That means, the decrease of yield strength due to the climb motion should be strongly dependent on the shape of dispersed particles.

Based on the local climb mechanism, Shewfelt and Brown$^{(4)}$ theoretically derived an equation for the yield strength of DH alloys. They examined the shape effect of dispersed particles on the yield stress by comparing their theoretical results with the experimental ones found in the literature of copper alloys with ill-defined platelets of Al$_2$O$_3$ or BeO and with spherical SiO$_2$ particles. They, however, could not confirm a clear shape effect$^{(4)}$. It is still unclear whether the absence of clear shape effect comes from some defect in the theory or from some experimental error (e.g., an inaccurate determination of the yield stress).

The following question remains unanswered in their analysis: If the dispersed particles in the specimens quoted by them were too small to have a detectable shape effect on the local climb (diameter; 56–81 nm), or if the shape effect was weaker than the effect of the dispersion parameter (particle size and interparticle distance) or the character of particles, or both the effects, because the dispersion parameter and the particle species were different between the reference specimens.

If the diminution of the yield strength at high temperatures occurs really through the climb of dislocations, it may be prevented by making the shape of particles complex, as shown in Fig. 1(c). In this case, dislocations are forced to pass the particles by the Orowan mechanism, because the dislocations encountered by particles will fall on the concave positions of the particles and can not climb over them.

In order to reveal the shape effect, it is advantageous to use the specimens with particles of the same species and dispersion parameter but largely different in shape. Internally oxidized Ni–Si alloys reported earlier$^{(9)}$ are a good candidate for this purpose, in which the shape of SiO$_2$ particles as oxidized is very complex like a dendrite, and it can be easily changed to spherical one without any change in the spacial distribution by a heat treatment at a higher temperature than the oxidation temperature.

In the previous paper$^{(9)}$, the shape effect of dispersed particles on the high-temperature hardness was investigated by using two kinds of nickel alloys with SiO$_2$ particles of complex and spherical shapes. However, against expectation, the shape effect on the temperature dependence of hardness was not detected even by a long loading time test performed to eliminate the work hardening effect. As pointed out previously$^{(9)}$, there were the following possibilities for the absence of the shape effect: the eliminating treatment of the work hardening effect was insufficient and the hardness test was not sensitive enough to reveal the shape effect.

In this paper, the shape effect is investigated on the high-temperature yield strength in the two kinds of specimens mentioned above by means of tensile test, which is thought more sensitive than the hardness test. For comparison, pure nickel specimens are also tested. Based on the experimental results, the shape effect of dispersed particles on the yield strength is quantitatively discussed.

II. Specimens and Experimental Procedure

1. Specimen preparation
A Ni–0.32 mass% Si alloy was melted and cast by using the same method as mentioned in the previous paper$^{(9)}$. After a homogenizing
heat treatment at 1373 K for $1.13 \times 10^5$ s (32 h) under a vacuum of $2.67 \times 10^{-4}$ Pa, the ingot was cold-rolled to about 1.0 mm in thickness (reduction; 90%) and then machined to specimens of $60 \times 10 \times 1.0$ mm.

After annealing at 1473 K for 57.6 ks (16 h), the specimens were internally oxidized at 1373 K for 468 ks, which was enough to oxidize Si to the center of the specimen (9). The grain size after the annealing was about 1.0 mm in diameter. The method of internal oxidation was the same as that mentioned in the previous paper (9). The microscopic observation on the section confirmed that the particles dispersed up to the center of specimens.

Half of the specimens were reserved as the as-externally oxidized specimens, which contained SiO$_2$ particles of complex shape (they will be referred to as Ni-Comp.SiO$_2$). The other half were heat-treated for 173 ks at 1573 K to spheroidize the particles (they will be referred to as Ni-Sphe.SiO$_2$). Then, both the specimens of Ni-Comp.SiO$_2$ and Ni-Sphe.SiO$_2$ were encapsulated in a fused silica tube with vanadium flakes (deoxidizer) under a vacuum of $2.67 \times 10^{-4}$ Pa and heated for 113 ks at 1573 K to remove oxygen resolved in the matrix. Vanadium is a favorable metal for removing oxygen resolved in the matrix without any reduction of SiO$_2$ particles, because the formation free energy of SiO$_2$ is larger than that of VO.

As well known, the size and interparticle spacing of dispersed particles in internally oxidized alloys tend to increase with the distance from the surface. Then, the mechanical properties may also change, depending on the distance from the surface. However, the change in the dispersion parameter is usually large only near the surface but relatively small in the interior region (10). Therefore, the inhomogeneously dispersed layer with very fine particles near the surface, which was about 0.1 mm thick in both the specimens of Ni-Sphe.SiO$_2$ and Ni-Comp.SiO$_2$, was removed by polishing with emery paper (#1500). By using a spark cutting method these specimens were machined and finished to tensile specimens with a gage part of $12 \times 3.5 \times 0.8$ mm and the shoulder curvature 3.0 mm in radius (60 mm in total length). The work hardening supposedly introduced by the fabrication was removed by an annealing for 1.8 ks at 1300 K before the tensile test.

In order to identify the shape of the dispersed particles, the sections of these specimens were deeply etched and examined by a scanning electron microscope. Micrographs obtained are shown in Fig. 2. Figure 2(a), which was obtained on the specimen deoxidation-treated after the internal oxidation, shows that the shape of each particle is very complex like a dendrite. On the other hand, Fig. 2(b), which was obtained on the specimen deoxidation-treated after the spheroidization, shows that most of the particles are spherical.

By comparing the particle distributions in Figs. 2(a) and (b), it is found that the particle dispersion (interparticle spacing) is not significantly changed by the spheroidizing treat-
ment, though the shape of the particles changes remarkably from dendritic to spherical. Therefore, when the particles are cut by an arbitrary plane, the apparent size of the individual particles on the plane should be smaller in Ni-Comp.SiO₂ than in Ni-Sphe.SiO₂, and in parallel the number density on the plane should be larger in the former, as expected by comparing Figs. 2(a) and (b).

Figures 3(a) and (b) are optical micrographs of unetched sections of the two materials, showing an actual evidence of the expected difference mentioned above. Therefore, the number of particles acting as obstacles to dislocation motion is much larger for the dendritic particles than for the spherical ones. Fine particles observed in Fig. 2(a) are not randomly distributed but tend to lie in lines in some directions, reflecting the dendritic form of particles.

By assuming random distribution, the mean particle diameter and the mean interparticle spacing were determined on an arbitrary section and are shown in Table 1 with the chemical composition of silicon. Here, the volume fraction of SiO₂ particles \( f_v \) was calculated on the assumption that all the silicon atoms in the matrix were oxidized.

| Specimen                  | Si (mass%) | \( f_v \) (%) | \( l_5 (\mu m) \) | \( L_5 (\mu m) \) |
|---------------------------|------------|----------------|-------------------|-------------------|
| Ni-Comp.SiO₂              | 0.324      | 2.72           | 0.64              | 4.16              |
| Ni-Sphe.SiO₂              | 0.324      | 2.72           | 0.88              | 5.84              |

\( f_v \): Volume fraction of SiO₂.
\( l_5 \): Mean particle diameter on the observed section.
\( L_5 \): Center to center interparticle spacing on the observed section.

Although it had been an anxious matter that the shape of dispersed particles would change during the tensile test at high temperatures, actually no change in the shape was observed after the test. This may be because that the temperature and time for the internal oxidation and the spheroidization were sufficiently higher and longer than those for the tensile tests.

Fig. 3 Optical micrographs showing the distribution of silica particles on a cross section; (a) fine distribution (on Ni-Comp.SiO₂) and (b) coarse distribution (on Ni-Sphe.SiO₂).

2. High temperature tensile test

Two series of tensile tests were conducted in a vacuum of \( 4 \times 10^{-3} \) Pa. One series was at a strain rate of \( 5.6 \times 10^{-5} \) s\(^{-1} \) at various temperatures from room temperature to 1273 K. The other series was at 1173 K at various strain rates from \( 2.58 \times 10^{-3} \) to \( 2.62 \times 10^{-5} \) s\(^{-1} \).

In order to detect the effect of climb motion of dislocations, the stress at the very beginning of plastic deformation must be exactly determined to avoid the confusing effect of work hardening\(^9\). Even when a precise stress-strain curve is obtained, a large discrepancy may occur depending on the definition of yield stress. For example, 0.2% offset stress, which is usually adopted as yield stress, is not always reasonable, because the work hardening introduced by the 0.2% strain depends on the material and testing condition, and in some cases the hardening is significant and in other cases insignificant.

In the present investigation, a load-displace-
ment curve was recorded with high chart speeds (1.66×10⁻³ ~ 3.33×10⁻² m/s) for constant cross head speeds (3.33×10⁻⁷ ~ 3.3×10⁻⁵ m/s) in order to enlarge the scale of the strain axis and clarify the initial part of the stress-strain curve. The yield stress was defined as the first deviating stress from the elastic linearity (σₚₑₙₚ), at which the plastic strain was as small as about 1×10⁻⁵.

It should be noted that in the case of testing machine system with a large elastic displacement the very beginning stress for plastic deformation can be hardly determined because the elastic displacement is much larger than the plastic one at the yield stress. Accordingly a specially high rigid machine was used in this investigation (spring constant kₘₛₚₑₚ = 58 MN m⁻¹). The machine⁴¹ was based on Shimazu Servo Pulser EHF-2 type. A typical example of the load-displacement curve is shown in Fig. 4, in which the point indicated by an arrow shows the yield stress defined above. Some reading error is also unavoidable, which depends on the drawing way of elastic linearity. However, the reading scatter of several times on the same load-displacement curve was as small as ±3%.

On the other hand, for the strain rate dependence of the yield stress, the true plastic strain rate must be also determined, because the displacement of the cross head includes the elastic displacement of the machine assembly. The following relationship holds between the true plastic strain rate, ̇εₚ, and the apparent strain rate, ̇εₚₐ, which includes the elastic strain¹¹.

\[
\dot{\varepsilon}_p = \dot{\varepsilon}_a \left(1 - \frac{1}{K \varepsilon_a} \right),
\]

where K is the apparent Young’s modulus or combined machine stiffness, and dσ/deₚₐ the apparent work hardening rate. The modulus K depends on the Young’s modulus and shape (gage length and cross section) of the specimen, and it ranged from 21.5 to 31 GPa in this experiment. The true plastic strain rate at the yield point was determined by using eq. (1).

### III. Experimental Results

#### 1. Stress-strain curve

Some typical examples of the stress-strain curves obtained in this experiment are shown in Fig. 5.

Figure 5(a) shows the true stress-true strain curves for pure nickel, Ni-Comp.SiO₂ and Ni-Sphe.SiO₂ specimens at 300 K at a strain rate of 5.6×10⁻⁵ s⁻¹, where each yield point is shown by an arrow. It is found that for pure nickel the yield stress and work hardening rate are low, while Ni-Comp.SiO₂ and Ni-Sphe.SiO₂ show very high yield stress and rapid work hardening nearly parabolic. These behaviors are characteristic of DH alloys. Further, both of the yield stress and the work hardening rate after yielding are always higher in Ni-Comp.SiO₂ than in Ni-Sphe.SiO₂, the difference in the shape of dispersed particles reflecting on the whole stress-strain curve.

Figure 5(b) shows the results at a higher temperature, 1073 K. The differences among the three materials are similar to those at 300 K (Fig. 5(a)), except that the stress level and the amount of work hardening are smaller than

![Fig. 4 A typical load-displacement curve for the Ni-Comp.SiO₂ specimen deformed at 873 K.](image-url)
those at 300 K. However, the shape effect on the amount of work hardening up to the steady state deformation (where the flow stress is almost independent of strain) almost disappears at 1073 K. Rather the work hardening is slightly larger in Ni-Sphe.SiO$_2$ than Ni-Comp.SiO$_2$. From the above results it is known that the shape effect on the stress-strain curve is remarkable at room temperature, and the more complex the shape of particles is, the more strongly the yield stress and work hardening rate increase, but the shape effect on work hardening becomes insignificant at high temperatures and a slight inverse effect happens to occur on the amount of work hardening. This inverse effect is clearly seen at higher temperature and lower strain rate (Fig. 6(b)).

Figure 6 shows the effect of strain rate on the shape of stress-strain curve. Figure 6(a) and (b) are the true stress-true strain curves obtained under the initial strain rates of $2.58 \times 10^{-3}$ s$^{-1}$ and $2.62 \times 10^{-5}$ s$^{-1}$, respectively. It is seen that the yield stresses in pure nickel and Ni-Comp.SiO$_2$ are almost unchanged by the change in strain rate by two orders of magnitude. On the other hand, the yield stress in Ni-Sphe.SiO$_2$ decreases with the decrease in strain rate. The amount of work hardening, however, decreases along with the strain rate in all the specimens. Further, at the high temperature and at the low strain rate shown in Fig. 6(b), the amount of work hardening up to the steady-state deformation is distinctly larger in the specimen with spherical particles than in the specimen with complex-shaped ones.

2. Temperature and strain-rate dependences of the yield stress

Figure 7 shows a plot of the yield stress, $\sigma_{p,1}$, against the testing temperature, $T$, for pure nickel, Ni-Sphe.SiO$_2$ and Ni-Comp.SiO$_2$. The yield stress is higher in the order of Ni-Comp. SiO$_2$, Ni-Sphe.SiO$_2$ and pure nickel in the whole temperature range and it tends to decrease as the temperature rises in each material. The decrement is especially large in Ni-Sphe.SiO$_2$. This fact shows that the more complex the shape is, the more effectively the particles increase the high-temperature strength.

Figure 8 shows the strain-rate dependence of yield stress in these materials at 1173 K, where the bulk-diffusion is high enough to allow the climb motion of dislocations. The yield stress
is higher in the order of Ni–Comp.SiO₂, Ni–Sphe.SiO₂ and pure nickel in the whole strain range. The strain-rate dependence is not observed in pure nickel and Ni–Comp.SiO₂ but observed in Ni–Sphe.SiO₂; the yield stress of Ni–Sphe.SiO₂ increases with the increase in the strain rate.

### IV. Discussion

#### 1. Shape effect of dispersed particles

It is known that the yield stress of DH alloys is approximately given by

\[ \sigma_y = \sigma_M + \sigma_p \]  

where \( \sigma_M \) is the yield stress of the matrix and \( \sigma_p \)
the contribution of dispersed particles.

In the case in which the Orowan mechanism operates, \( \sigma_p \) is given by the Orowan stress, \( \sigma_{or} \),

\[
\sigma_p = \sigma_{or} = 3.06 \frac{0.81Gb}{2\pi(L_i - l_i) \sqrt{1-\nu}} \ln \left( \frac{L_i}{b} \right),
\]

where \( L_i \) is the mean planar interparticle spacing, \( l_i \) the mean planar particle diameter and \( \nu \) the Poisson's ratio \(^{12} \). Therefore, the modulus-compensated \( \sigma_p, (\sigma_p - \sigma_M)/G = \sigma_p/1.\), should be a constant independent of temperature and strain rate. On the other hand, when the climb mechanism operates, \( \sigma_p \) should decrease as the temperature rises or the strain rate decreases.

From the results shown in Figs. 7 and 8, the increment of modulus-compensated yield stress, \( \Delta \sigma_{p,1} G_{RT}/G_T \), where \( G_{RT} \) and \( G_T \) are the shear moduli at room and test temperatures, respectively, is calculated and shown in Figs. 9 and 10. The Orowan stress, which was calculated by substituting the values of \( L_s \) and \( l_s \) listed in Table 1 for eq. (3), and the threshold stress for the local climb mechanism, which was estimated by the theory of Shewfelt and Brown \(^6\), are also shown in these figures, where the Orowan stress is denoted by \( \sigma_{or} \) for Ni-Comp.SiO\(_2\) and \( \sigma_{or} \) for Ni-Sphe.SiO\(_2\), and the threshold stress by \( 0.4 \sigma_{or} \) for Ni-Comp.SiO\(_2\) and \( 0.4 \sigma_{or} \) for Ni-Sphe.SiO\(_2\).

As shown in Fig. 3(a), in a specimen with particles of complicate shape, the particles on an arbitrary plane, say a slip plane, tend to cluster in lines. The dislocations may preferentially pass through the region of large interparticle spacings. Then the mean distance between these clusters on optical micrographs was measured and the Orowan stress was estimated from the distance. The estimated value is also shown in these figures as \( \sigma_{or} \) (loc. dist.).

As seen in Fig. 9, the value of \( \Delta \sigma_{p,1} G_{RT}/G_T \) for Ni-Comp.SiO\(_2\) is close to the value of \( \sigma_{or} \) independent of temperature. Therefore, it is presumed that the local climb mechanism does not operate for the complex-shaped particles and \( \sigma_p \) is determined by the Orowan mechanism.

On the other hand, the \( \Delta \sigma_{p,1} G_{RT}/G_T \) for Ni-Sphe.SiO\(_2\) decreases at high temperatures; it is nearly equal to \( \sigma_{or} \) at room temperature but decreases to about 0.45 \( \sigma_{or} \) at the highest temperature tested. Then, it is presumed that the yield stress for Ni-Sphe.SiO\(_2\) is determined by the Orowan mechanism at room temperature but determined by the climb mechanism at high temperatures.

From the evidence mentioned above, it is concluded that the shape effect of dispersed particles really exists in SiO\(_2\) dispersed nickel alloys; the specimen with spherical particles is deformed by the climb mechanism at high temperatures but the specimen with complex-shaped ones by the Orowan mechanism even at high temperatures. According to the fact that
the yield stress for both the DH nickel alloys does not depend at room temperature on the shape of dispersed particles and agrees with their Orowan stress, the difference in the yield stress at room temperature is understood to be caused by the difference in the particle number density on the slip plane. That is, it is another shape effect that the complex-shaped particles increases the Orowan stress by increase in the density of particles on the slip plane.

The fact that the yield stress increment in Ni–Sphe.SiO₂ decreases to about one fourth of that in Ni–Comp.SiO₂ at 1273 K, where the diffusion rate is presumably high enough to allow the climb mechanism to operate, is considered to come from the two shape effects mentioned above.

Figure 10 shows the strain rate dependence of \( \Delta \sigma_{p.l} G_{RT}/G_T \) at 1173 K. It is found that the value of \( \Delta \sigma_{p.l} G_{RT}/G_T \) for Ni–Sphe.SiO₂ depends on the strain rate and increases along with strain rate in a range between \( \sigma_{or}^S \) and \( 0.4 \sigma_{or}^S \), though the value for Ni–Comp.SiO₂ is independent of the strain rate and close to the Orowan stress, \( \sigma_C^O \) or \( \sigma_C^{O(loca. dist.)} \). The strain rate dependence also supports the understanding that the yield stress for the specimen with complex-shaped particles is determined by the Orowan mechanism, but the specimen with spherical ones by the climb mechanism.

There is a tendency that the amount of work hardening up to the steady-state deformation is larger in Ni–Sphe.SiO₂ than in Ni–Comp.SiO₂ at high temperature, as shown in Fig. 6(b). This fact may be explained as follows; since both of the DH nickel alloys have almost the same spacial distribution of dispersed particles, most of the shape effect on their strength may disappear after the formation of dislocation tangles around each particle, as mentioned in the previous paper. Then, the amount of work hardening, which is given by the difference between the stresses at the yield point and at the steady-state deformation, becomes larger in the specimen with spherical particles than that with complex-shaped particles.

Contrary to the present evidence of the shape effect on the temperature and strain-rate dependences of the yield stress obtained by the tensile test, the shape effect was not clearly observed by the micro-Vickers hardness test, as mentioned in the previous paper. The discrepancy may arise from a strong effect of work hardening included in the hardness test.

2. Mechanism for climb motion of dislocations

As shown in Figs. 9 and 10, the climbing of dislocations over the particles was considered to occur in the specimen with spherical particles. As mentioned in section I, the local and general climb mechanisms have been considered so far. Although several theories have been proposed for the local climb mechanism, we will analyse the date according to the Shewfelt and Brown’s model, because their model well explains the present results on the specimen with spherical particles but the other models do not.

Shewfelt and Brown have derived the following relation among \( \Delta \sigma_{p.l} \), \( T \) and \( \dot{\varepsilon} \) by assuming the local climb, where the moving dislocations locally pass over the particular obstacles by the aid of vacancy (or atom) flow induced by the concentration gradient of vacancies around the interface,

\[
\Delta \sigma_{p.l} \frac{G_{RT}}{G_T} = 3.06 \frac{G_{RT} b}{L_s} \left[ 0.51 + 0.12 \log \left( \frac{3.06 k T R^2 \dot{\varepsilon}}{4 \pi \rho b^2 a G_T L_s D_0} \right) \right] + 0.052 \frac{Q_l}{k T},
\]

where \( k \) is the Boltzmann’s constant, \( R \) the particle radius, \( \rho \) the mobile dislocation density, \( a \) the climb area associated with a vacancy, \( D_0 \) the frequency factor of the lattice self-diffusion coefficient in the matrix and \( Q_l \) the activation energy for the self-diffusion.

Since the dislocation density at the yield point, \( \rho \), should not depend on the temperature and strain rate, eq. (4) may be rewritten as follow,

\[
\Delta \sigma_{p.l} \frac{G_{RT}}{G_T} = A + 0.159 \frac{G_{RT} b Q_l}{L_s} \frac{Q_l}{k T} + 0.36 \frac{G_{RT} b}{L_s} \log \dot{\varepsilon},
\]
where $A$ is a constant independent of temperature and strain rate. For the derivation of eq. (5), the temperature dependence of the second term in eq (4) is neglected compared with that of the third term, because the logarithmic dependence is much smaller than the inverse dependence. From eq. (5), a linear relationship holds between $\Delta \sigma_{\text{p,1}} G_{\text{RT}}/G_T$ and $1/T$ under constant $\dot{\varepsilon}$ or $\Delta \sigma_{\text{p,1}} G_{\text{RT}}/G_T$ and log $\dot{\varepsilon}$ at a given $T$.

Equations (4) and (5) hold only in the stress range of the Orowan stress, $\sigma_{\text{or}}$, to the threshold stress for yielding, $0.4 \sigma_{\text{or}}$, below which dislocation cannot pass over the particles.

As shown in Fig. 9, the lowest value of yield stress obtained for the specimen with spherical particles at 1273 K was about $0.45 \sigma_{\text{or}}^5$, which was very close to the lowest threshold stress, $0.4 \sigma_{\text{or}}$, estimated by Shewfelt and Brown. Then, the range in our experimental conditions satisfies the holding condition for eq. (4).

In order to examine the validity of eq. (5), the data for Ni–Sphe.SiO$_2$ in Fig. 9 was replotted as $\Delta \sigma_{\text{p,1}} G_{\text{RT}}/G_T$ vs. $1/T$ in Fig. 11. As expected from eq. (5), the datum points show roughly a linear relationship in a high temperature range from 873 K to 1273 K, where the diffusion rate may be high enough to allow the climb motion of dislocations, but not in a low temperature range below 873 K. The slope of this line is 15.2 GPa K$^{-1}$. On the other hand, substitution for eq. (5) of the values of $b=0.25$ nm, $L_s=4.96 \mu$m, $Q_L=281$ kJ/mol, $G_{\text{RT}}=83.84$ GPa$^{15}$ and $G_T^{16}$, gives a slope of 22.8 GPa K as an average numerical coefficient of $1/T$ in the second term on the right hand side of eq. (5) in the high temperature range. The calculated slope is fairly close to the experimental value mentioned above.

Next, we will discuss the strain-rate dependence of $\Delta \sigma_{\text{p,1}} G_{\text{RT}}/G_T$. As seen in Fig. 10, a good linear relationship between $\Delta \sigma_{\text{p,1}} G_{\text{RT}}/G_T$ and log $\dot{\varepsilon}$ also holds in the yield stress range from $\sigma_{\text{or}}^5$ to $0.4 \sigma_{\text{or}}$ for Ni–Sphe.SiO$_2$. The slope of this relationship is 3.16 MPa. The coefficient of log $\dot{\varepsilon}$ in the third term on the right hand side of eq. (5) is calculated to be 4.03 MPa by substituting for eq. (5) the values of $G_{\text{RT}}$, $b$ and $L_s$ mentioned above. The calculated slope is very close to the value obtained in this experiment.

From the quantitative analysis, it is further found that these values mentioned above are inconsistent with any other theoretical equations so far proposed$^{13,14}$. Therefore, the yield stress of Ni–Sphe.SiO$_2$ is assumed to be determined by the local climb mechanism proposed by Shewfelt and Brown$^{4}$.

On the other hand, the general climb mechanism$^{6,6}$ has been also proposed as another mechanism for the climb motion of dislocations. According to this mechanism, dislocations pass over the particulate obstacles by means of the climbing of the whole dislocation length between the dispersed particles after touching the particle/matrix interface, and the increment of dislocation length during the climb over the particles is much smaller than that in the local climb mechanism. That means, the general climb can occur under a much smaller stress than that for the local climb, but it requires a large amount of vacancy flow. Therefore, the general climb mechanism is characterized by the fact that it operates only under the conditions of low strain rate and high temperature. In fact, the value of only about one tenth of the lowest yield stress estimated by the local climb mechanism$^{6}$ is predicted as the threshold stress by the general climb mechanism. Only one hundredth of the strain rate for the local climb mechanism is required for the general climb$^{6}$.

Since the present experimental results are in relatively good agreement with the predictions...
of the local climb mechanism, the general climb mechanism is hardly considered to operate under the present experimental conditions. This mechanism may occur under the conditions of much higher temperature or much lower strain rate than those in the present experiments.

V. Conclusion

In order to investigate the shape effect of dispersed particles on the yield strength, two kinds of nickel alloys with SiO₂ particles of complex and spherical shapes were made by internal oxidation and a heat-treatment after the internal oxidation, respectively. Their yield stresses were measured by means of tensile test in a temperature range from 300 K to 1273 K at a strain rate of 5.6 × 10⁻³ s⁻¹, and in a strain-rate range from 2.67 × 10⁻³ s⁻¹ to 2.62 × 10⁻⁵ s⁻¹ at 1173 K. The results are as follows.

(1) In the specimen with complex-shaped particles, the increment in modulus-compensated yield stress due to the particles depends neither on the temperature nor on the strain rate and agrees with the Orowan stress. This fact suggests that the climb of dislocations over the particles does not occur in this specimen.

(2) In the specimen with spherical particles, the yield stress increment due to the particles agrees with the Orowan stress at room temperature but decreases with the increase in temperature. Further, the yield stress increment at 1173 K is proportional to the logarithm of the strain rate.

(3) By the analysis of temperature and strain rate dependences of the yield stress, it is confirmed that the yield stress increment of the specimen with spherical particles was controlled by the local climb mechanism proposed by Shewfelt and Brown.

(4) The complex-shaped particles effectively raise the Orowan stress with increase in the particle number density on the slip plane.

(5) The yield stress in the specimen with spherical particles is always lower than that with complex-shaped particles, because of the effects of the climb motion of dislocations in addition to the smaller number density of dispersed particles on the slip plane.

(6) It is concluded that making the shape of dispersed particles complex is quite advantageous for the improvement of high temperature strength of dispersion-hardened alloys.

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