EFFECT OF HEAT TREATMENT ON MICROSTRUCTURE AND TENSILE DEFORMATION CHARACTERISTICS

OF UDIMET 520

H. Doi and Y. Shimanuki
Central Research Laboratory of Mitsubishi Metal Corporation, Omiya City, Saitama Prefecture, Japan

Effect of age-hardening treatment of Udimet 520 on growth of γ' particles and the correlation between the resulting microstructure and tensile deformation characteristics were investigated in some detail. The observed variations of the room-temperature proof stress, strain-hardening coefficient and internal friction with temperature and time of the age-hardening treatment can be explained if we take into account of the effect of transition in the mode of dislocation motion from shearing to by-passing as a result of coarsening of γ' particles. A quantitative analysis of the age-hardening of the alloy on the basis of the theoretical model by Brown et al. of shearing of γ' particles by paired dislocations revealed that modification of the existing theory is necessary. Two types of serrated yieldings were found by elevated-temperature tensile testing: the analysis of the effect of temperature and strain rate suggests that one type of serration (referred to as Type A) which appears in the temperature range of 200 to 350°C corresponds to the dynamic strain aging due to Al and Ti atoms and another (Type B) which appears in the temperature range of 350 to 600°C to repeated shearing of γ' particles by moving dislocations. Analysis of the dependence of the elevated-temperature proof stress on the volume fraction of γ' phase and direct observations of γ'-dislocation interaction revealed that when the specimen is deformed at temperatures below 700°C, γ' particles of comparatively small sizes are invariably sheared by moving dislocations.

Introduction

Ni-base superalloys are characterized by the basic structure of Ni-Cr matrix precipitation hardened by γ' (Ni_3 (Al, Ti)). However, they are usually further hardened by variation of relative amounts of Al and Ti, or by some optimum additions of such elements as Co, Fe, Mo, W, Nb, B, Zr, C, etc. and as a result, various grades of superalloys have evolved (1). It is to be noted that the evolution of these grades has been the result of empirical search; so are the optimum heat treatment condition, now in wide practice, which has been determined to realize optimum strength properties at high operating temperatures. Although a number of basic studies on some simple binary or ternary alloys have been reported (2), with reference to commercial superalloys containing a number of alloying elements, systematic analysis of the effect of temperature and time of heat treatment on the variation of the microstructure and mechanical properties has hitherto been comparatively scarce.

The strengthening mechanisms of superalloys are beginning to be understood as a result of the recent worldwide research (3). The hardening is believed to be determined by the relative contribution of the following factors: (a) anti-phase boundary hardening due to ordered γ', (b) coherency strain hardening due to γ'-γ mismatch, (c) hardening due to the carbide precipitation on grain boundaries, (d) solid solution hardening of the matrix.

Strengthening by γ' particles has been investigated on some simple Ni-base alloys and several theoretical expressions for the increment in room-temperature proof stress have been reported (2-4). In regard to the elevated-temperature mechanical properties, the effect of thermal activation on dislocation movements (2) and the effect of variation with temperature of γ' strength on the proof stress have been studied (5).

Solid-solution hardening capability of various alloying elements up to 0.6Tm has been related to atomic size misfit as measured by crystal lattice expansion and to lowering of stacking fault energy, etc. (3). Above 0.6Tm, the slow diffusing elements such as Mo and W are believed to be most potent solid-solution hardeners since in such high temperature range strengthening is believed to be diffusion controlled (3).

However, it seems that various aspects of the effect of heat treatment on strengthening of superalloys have by no means completely understood. In particular, the accumulation of experimental data in regard to the systematic analysis of deformation characteristics of commercial superalloys as a function of coarsening of precipitated γ' is necessary.

In the present study, as the commercial superalloys suitable for the investigation of the effect of heat treatment, Udimet 520 was chosen for its desirable microstructure; precipitated γ' particles are in general spherical and uniformly distributed. The pertinent properties investigated are tensile deformation characteristics; particular attention is directed toward correlation between γ' coarsening due to aging at some appropriate temperatures and the resulting variation in the deformation characteristics.

Experimental Procedure

Udimet 520 alloys A and B with the composition shown in Table 1 were used in the present experiment. Specimens were machined from thin sheets, 1.5 mm thick, which were fabricated by the conventional process involving vacuum melting, hot forging and hot rolling. Solution treatment of the specimens was invariably carried out by heating for 4 hours at 1120°C followed by water quenching. Aging treatment was carried out by heating at 700 and 800°C for various lengths of time. Some of the specimens were subjected to the conventional heat treatment established for Udimet 520, i.e. 1120°C x 4 hrs. sol. tr. $\longrightarrow$ A.C. $\longrightarrow$ 843°C x 24 hrs. aging $\longrightarrow$ A.C. $\longrightarrow$ 760°C x 16 hrs. aging $\longrightarrow$ A.C. The heat treatment was invariably carried out by sealing the specimen in a quartz tube in vacuum of 5×10^{-5} torr. Grain-size measurements of the specimens were carried out by the standard method; it was calculated to be $\sim$0.06 mm.

Room-temperature tensile tests of the specimen of 1.5 mm x 6 mm cross section and 60 mm gauge length were carried out using the Instron testing machine at a strain rate of 0.7% per minute. Tensile tests at elevated-temperatures were carried out *in vacuo* with the specimen of 1.3 mm x 2 mm cross section and 17 mm gauge length at a strain rate of 2% per minute; for that purpose a resistance-heating electric furnace was attached to the Instron testing machine. Internal friction measurements were carried out with age hardened and lightly deformed specimens of 2 mm x 6 mm x 100 mm in size; the flexural vibration at a resonant freqnency of 1 kHz was utilized. The measurements were invariably carried out *in vacuo* at the maximum strain amplitude of 1×10^{-5} which was in the strain amplitude independent region for the specimen. Transmission electron-microscopic observations were carried out using JEM 120 electron microscope. Thin foils were prepared from rolled sheets of 0.1 mm thickness by electropolishing in a electrolyte of 10% $HClO_4$ + ethanol at the current density of 300-400 mA/cm^2.

Table 1. Chemical Composition of Specimens Investigated

	Cr	Co	Mo	W	Ti	Al	C	B	Ni
Alloy A	18.5	12.5	6.2	0.9	3.0	2.2	0.03	0.006	Bal.
Alloy B	18.5	12.1	6.4	0.3	3.2	1.9	0.03	0.003	Bal.

Results

Effect of Age-Hardening Treatment on Room-Temperature Tensile Properties

Variation with aging time at 700 and 800°C of 0.2% proof stress and the mean strain-hardening coefficient of alloy A at the strain interval of 2 and 3% was shown in Fig. 1. In the case of 800°C aging, the proof stress rises rapidly with time and reaches to the peak value by aging for as short as 3 hours. On the other hand, when the temperature for age hardening is lowered to 700°C, the proof stress does not reach to the peak value even after aging for 100 hours; moreover the peak value appears to be much higher than that reached by aging at 800°C. In regard to the strain-hardening coefficient, it is to be noted that it remains constant for some length of time in the early stage of age hardening and after some critical time which depends on the aging temperature it rises rather rapidly with time; the strain-hardening coefficient begins to rise prior to the proof stress reaching to the peak value.

In order to obtain information on the relative contribution to the proof stress of grain interior and grain boundaries, specimens of grain diameters of 0.02 to 0.2 mm were prepared by varying the heating time while the solution-treatment temperature was fixed as 1200°C. The values of the proof stress obtained from tensile tests were then correlated to the grain diameter D by the well-known Hall-Petch relation (6):

$$\sigma_{0.2} = \sigma_0 + kD^{-\frac{1}{2}} \tag{1}$$

In Fig. 2, the relative contribution of these two terms, σ_0 and $kD^{-\frac{1}{2}}$, to the proof stress was shown against aging time. It can be seen that σ_0 markedly increases with age-hardening treatment evidently owing to the precipitation of γ' particles and this occupies a major fraction of the proof stress of the alloy.

Variation of Internal Friction with Aging Time

In order to obtain information on the interaction between precipitated γ' particles and moving dislocations, measurements of internal friction of the specimen were carried out as a function of aging time. In Fig. 3, variation with aging time of internal friction of alloy A is shown; the lowest curve corresponds to internal friction of the as aged specimen whereas the upper two curves to that of the age-hardened and 3% deformed specimens. Internal friction of the specimen age-hardened at 800°C was found to be rather low (5×10^{-5}) and decreases monotonically with the aging time; a similar result was obtained with specimens age-hardened at 700°C. A possible interpretation for this observation is that precipitated γ' particles act as pinning obstacles for the oscillatory motion of dislocation segments and thereby lowers the amplitude independent component of internal friction. On the other hand, internal friction of the age-hardened and deformed specimens rises at some critical aging time which in general depends on temperature. Considerations of the nature of dislocation damping and the variation in the microstructure with aging time which will be described in the next section suggest that this is related to a change of the deformation-induced dislocation configuration with coarsening of precipitated γ' particles. In this connection, it is interesting to note that the increase of internal friction of the age-hardened and deformed specimens resembles rather closely to the increase of the strain-hardening coefficient; both start to rise at the same aging time prior to the proof stress reaching to the peak value (see Fig. 1).

Variation of Microstructure with Aging Time

Growth of γ' particles by age-hardening treatment was investigated by means of transmission electron microscopy. Fig. 4(a), (b) and (c) show the dark-field images of γ'

particles of the specimens preliminarily aged for 0.5, 3 and 50 hours at 800°C, respectively. From such micrographs, the mean particle diameters were calculated to be 80, 180 and 520 Å corresponding to 0.5, 3 and 50 hours aging, respectively. It can be seen that the shape of γ' particles of Udimet 520 remains spherical even in the overaged condition. This can be attributed to the comparatively small γ'-γ mismatch; determination of the lattice constant of the extracted γ' particles in the age-hardened alloy by means of X-ray diffraction technique revealed that the mismatch remains as small as 0.2%.

Fig. 4(d) and (e) show the dislocation structure in the age-hardened and lightly deformed specimens. Existance of the array of paired dislocations in the specimen aged for only 0.5 hours at 800°C (Fig. 4(d)) indicates that dislocations move by shearing γ' particles, whereas dislocation loops are frequently observed in the specimen in the overaged condition, i.e. aged for 50 hours at the same temperature (Fig. 4(e)). Such dislocation loops are believed to have been formed in the process of dislocation by-passing; this means that dislocation can not shear coarse γ' particles in the overaged condition. More detailed observation revealed that dislocation loops and bowing-out are visible in the room-temperature deformed specimen which has preliminarily been aged for more than 3 hours at 800°C; the mean γ' diameters for the specimens subjected to such age-hardening treatments were found to be 180 Å. It should be noted that paired dislocations coexist with dislocation loops in the overaged specimen as can be seen in Fig. 4(e); this means that some fraction of γ' particles of comparatively small sizes are sheared by moving dislocations.

It is interesting to note that aging for 3 hours at 800°C gives rise to prominent variation in several properties; for example the proof stress reaches to the peak value and the strain-hardening coefficient as well as internal friction of the age-hardened and deformed specimen markedly rises by this heat treatment. As discussed in more detail later, these changes can consistently be explained on the basis of the variation in the mode of γ'-dislocation interaction.

Tensile Deformation Characteristics at Elevated Temperatures

Fig. 5 shows the stress-strain curves of alloy B obtained by tensile testing at R.T., 200, 500 and 800°C. The stress-strain curve for the room-temperature deformation is in general smooth and it has comparatively high strain-hardening coefficient, whereas serrated yieldings were found to take place with deformation at 200 and 500°C. The type of serrated yielding observed for deformation at 200°C considerably differs from that for deformation at 500°C. For convenience, the former is referred to as Type A and the latter as Type B serration. Type A serration has the following characteristics: (1) the temperature range for the occurrence is 200 to 350°C, (2) the serration is irregular and wavy in form and it first appears at the critical strain, referred to as ε_c, (3) it occurs with solution-treated as well as with age-hardened specimens; (4) the stress amplitude for each serration increases with strain, and (5) ε_c increases with strain rate and decreases with the rise in testing temperature. These characteristics are exactly the same as those of the Portvin-Le Chatelier effect which has frequently been observed with some fcc and bcc alloys (7). The P-L effect is interpreted to be due to the dynamic strain-aging and usually analysed on the basis of the following formula

$$\dot{\varepsilon} \propto \varepsilon_c^m \exp\left(-\frac{Q}{kT}\right) \qquad (2)$$

where m is a constant, k is the Boltzmann constant, and Q is the activation energy. The above relationship was applied to the case of Type A serration of Udimet 520; m and Q for the solution-treated specimen were calculated to be 2.11 and 0.85 eV, respectively. A serration similar to the Type A was also found to appear on the stress-strain curve of Ni-6.9%Al; m and Q for the alloy preliminarily aged for 0.5 hours at 700°C were found to be 2.12 and 0.91 eV, respectively.

On the other hand, Type B serration has the following characteristics: (1) the temperature range for the appearance is 350 to 600 °C, (2) it appears only with the age-hardened alloys, (3) it is characterized by the repeated yield drop of comparatively large amplitude and it appears with the onset of plastic flow; in some cases, the yield drop amounts to as much as ≥ 5 kg/mm^2, and (4) the stress amplitude for the serration increases with strain, aging time and testing temperature and decreases with strain rate. Type B serration also takes place with age-hardened Ni-6.9%Al.

In Fig. 6, the proof stress is plotted against testing temperature. The fully age-hardened specimen maintains its high room-temperature strength up to considerably high temperatures. The two-steps age-hardening treatment which has been established for Udimet 520 yields the highest strength up to 800°C. In the proof stress vs. testing temperature curve for the weakly age-hardened specimen, a hump appears in the temperature range from 600 to 800°C. Such a hump is supposed to have been formed as a result of the growth of γ' particles during the tensile test.

Microstructure of Alloys Preliminarily Age-Hardened and Deformed at Elevated Temperatures

Fig. 7(a) shows the array of dislocations for the alloy B specimen preliminarily aged for 0.5 hours at 800°C and deformed at 500°C. The existance of paired dislocations gives the evidence for shearing of γ' by dislocations as in the case for room-temperature deformation. Fig. 7(b) shows the dislocation tangles formed in the same specimen deformed at 900°C. It is to be noted that shearing of γ' by dislocations does not take place at this testing temperature; the mode of γ'-dislocation interaction is invariably by-passing of γ' particles. Microscopic observation of the specimen in a weakly age-hardened condition, i.e. aging at 800°C for less than 3 hours revealed that by deformation at temperatures below700°C, shearing of γ' by dislocations invariably takes place. On the other hand, by deformation of the specimen at temperatures above 700°C, dislocation loops have been formed on a different plane from the original one where dislocation by-passing was initiated; this suggests that dislocation loops have been formed by double cross slip of dislocations around γ' particles. When the time of age-hardening at 800°C exceeds 3 hours, the mode of γ'-dislocation interaction at the testing temperatures below 700°C is essencially the same with that observed in the specimen deformed at room temperature. In the case of deformation at the temperatures above 700°C, the mode of the interaction is similar to that observed in the weakly age-hardened specimen.

Discussion

Nature of Age-Hardening of Udimet 520

Contribution of γ' precipitation to age-hardening of Udimet 520 can be estimated from the result of Hall-Petch analysis shown in Fig. 2 if we assume that solid-solution hardening is not significantly affected by γ' precipitation. In the case of 800°C aging, the maximum contribution of γ' precipitation comprises as much as 50% of the highest proof stress value of the alloy. The precipitation of the carbides on the grain boundaries on the other hand plays only a minor role in the age-hardening of the alloy since its contribution amounts to only 7%. If we take into account of the fact that the matrix composition continuously changes during age-hardening, the contribution of γ' precipitation to the age-hardening of the alloy may somewhat increases.

Now, it has been demonstrated through transmission electron-microscopy of the alloy preliminarily age-hardened and lightly deformed at room temperature that the microstructure corresponding to the peak position on the age-hardening curve of Udimet 520 is characterized by the formation of the dislocation loops. Singhal et al. (8) reported a similar change in the mode of γ'-dislocation interaction in a Fe-Ni-Cr-Ti alloy.

An aspect which is worthy of attention is that shearing of γ' by dislocations has been observed even in the overaged condition (see Fig. 4(e)). Since γ' particles have their own size distribution and moreover, smaller particles are easily sheared by moving dislocations, it is understandable that the two modes of γ'-dislocation interaction, i.e. shearing and by-passing coexist even in the overaged condition.

It is to be noted that since by-passed dislocations usually form loops and tangles, dislocation density in an overaged specimen should more rapidly increase with plastic deformation than in an underaged specimen. This gives the most adequate explanation for the rapid rise of the strain-hardening coefficient after heat treatment which gives the maximum age-hardening; the rise in the strain-hardening coefficient can simply be attributed to the increase of dislocation-dislocation interactions. A similar observation on the increase of the strain-hardening coefficient has been reported by Parker (9) with Ni-Al-Ti alloys. The rapid rise of internal friction of preliminarily age-hardened and room-temperature deformed Udimet 520 (see Fig. 3) can also be sttributed to the rapid increase of dislocation density as a result of dislocation by-passing. Present experiment has revealed that internal friction technique can be a useful tool for sensitively detecting the change of dislocation configuration in a precipitation hardened alloy.

A question arises as to whether age-hardening of Udimet 520 due to γ' precipitation can satisfactorily be accounted for by the theoretical models of shearing of γ' by paired dislocations. Brown et al. (4) have derived the following theoretical expression for the stress increment due to precipitation of ordered particles by assuming that trailing dislocation is nearly straight:

$$\Delta\tau_\gamma = \frac{\gamma}{2b}\left(\frac{r_s}{R} - f\right) \tag{3}$$

where r_s and 2R are the mean planar particle radius and the mean particle separation along the leading dislocation, respectively, and f, γ and b are the volume fraction of γ' particles, the anti-phase boundary energy of γ' phase and the Burgers vector, respectively. Assuming that the particles arrange in a simple cubic array, R is expressed as

$$R = \frac{1}{2}\sqrt{\frac{\pi}{f}}\, r_s \tag{4}$$

and $\Delta\tau_\gamma$ is given by the following formula:

$$\Delta\tau_\gamma = \frac{\gamma}{2b}\left(2\sqrt{\frac{f}{\pi}} - f\right) \tag{5}$$

If we apply eq. (5) to the age-hardening of Udimet 520 (aged at 800°C for 1 hour) and use the measured value for f in eq. (5), we obtain the theoretical value for the increment in proof stress due to shearing of γ' as 20 kg/mm^2. Now, in order to obtain the value for the increment in proof stress, the contribution from the coherency strain-field should be taken into account. The calculation on the basis of the theory by Gerold (10) shows that for the case of Udimet 520 this contribution amounts to 5 kg/mm^2. By the addition of the calculated value for the effect of the coherency strain field to eq. (5), we obtain the theoretical value for the increment in the proof stress for Udimet 520 as 25 kg/mm^2, in rather marked discrepancy from the measured value of 35 kg/mm^2 (see Fig. 2). This large discrepancy is presumably due to the somewhat arbitrary assumption regarding the configuration of the trailing dislocation and to the simplicity in the assumption regarding the spatial distribution of γ' particles. It seems that a more refined theoretical treatment on the interaction between the paired dislocations and γ' particles and a more stringent comparison of the theoretical values for the proof stress and the measured values are required.

Nature of Serrated Yieldings of Udimet 520

As described in the previous section, the activation energy of the Type A serration for solution-treated Udimet 520, 0.85 eV, agrees well with that for weakly age-hardened Ni-6.9%Al, 0.91 eV. The value of the activation energy for the migration of Al in Ni can be estimated as 1.0 eV; this value is obtained by subtracting the formation energy of vacancy in Ni, 1.8 eV (11), from the activation energy for diffusion of Al in Ni, 2.8 eV (12). Since the value of the activation energy for the Type A serration for Ni-6.9%Al is nearly the same with that of migration of Al in Ni, the serration of Ni-6.9%Al can be attributed to the dynamic strain-aging by Al atoms. A comparison of activation energy for the Type A serration between Ni-6.9%Al and Udimet 520 suggests that Al atoms are also responsible for the dynamic strain-aging in Udimet 520.

Other possible solute elements which may cause the Type A serration are W, Mo, and Ti since the atomic oversizes are comparatively large, i.e. 13, 12 and 9%, respectively. Among them, Ti is considered to be the most important element; the activation energy for diffusion of Ti in Ni is approximately the same with that of Al in Ni and consequently its contribution to the dynamic strain-aging is supposed to be equivalent to that of Al. In comparison with Al and Ti, the contribution of Mo or W may be comparatively small since the activation energy for diffusion in Ni is much larger than that for Al or Ti.

The nature of the Type B serration can be clarified if we take into account the marked aging time dependence of stress amplitude for each yield drop. The large yield drop can be attributed to the interaction between dislocations and γ' particles; movements of many dislocations through repeated shearing of γ' on one slip plane decrease the total area of the anti-phase boundaries in that slip plane and this can lower the flow stress. Each yield drop of the Type B serration corresponds to some catastrophic movement of dislocations by shearing γ' particles. The dependence of the stress amplitude for each yield drop on the aging time can be explained by the growth of γ' particles; likewise the dependence on the testing temperature is explained from the positive temperature dependence of γ' strength.

Effect of Heat Treatment on Proof Stress at Elevated Temperatures

As shown in Fig. 6, variation with temperature of the proof stress of Udimet 520 depends on the time for the age-hardening treatment. Since the age-hardening condition governs the growth of γ' particles, it is of interest to investigate the relation between the variation with temperature of the proof stress and the volume fraction of γ' particles. The volume fraction of γ' in specimens shown in Fig. 6, preliminarily aged at 800°C for 0.5, 3 and 30 hours, have been estimated from the extraction method (13) as 20, 24 and 26%, respectively. Beardmore et al. (14) reported that temperature dependence of the proof stress of the Ni-base alloys which contain various amounts of γ' particles can be classified into three types according to the volume fraction of γ' phase: when the volume fraction of γ' is much larger than 50%, the proof stress of the alloy is characterized by the marked increase with deformation temperature (Type 1), whereas the proof stress of the alloy which contain comparatively small volume fraction of γ' particles (for example as small as 20%) decreases slightly with temperature (Type 2). In the intermediate range in the volume fraction of γ', the proof stress stays nearly constant with variation in temperature (Type 3). It is to be noted that temperature dependence of the proof stress of Udimet 520, Fig. 6, more or less resembles the Type 2; a comparison shows that the temperature dependence for age-hardened Udimet 520 is roughly the same with that of Ni-Cr-Al alloy which contains 20% γ' (14).

Although the volume fraction of γ' phase has not been widely varied in this study, a clear dependence of the elevated-temperature proof stress on the volume fraction has been shown; the proof stress of the specimen which contains larger volume fraction of γ' drops only slightly with temperature up to 800°C in comparison with the specimen which contains lower volume fraction of γ'. Observations by means of transmission electron-microscopy have revealed that the elevated-temperature proof stress is determined by the increase of the volume fraction of γ' particles sheared by dislocations even in the overaged condition, i.e. aging for more than 3 hours at 800°C although the probability for shearing of γ' in general decreases by coarsening of γ'.

Summary

The effect of temperature and time for the age-hardening of Udimet 520 on the growth of γ' particles and the tensile deformation characteristics was investigated through tensile tests, transmission microstructure observations and internal friction measurements. The results are summarized as follows.

(1) Hall-Petch plots of the proof stress for 800°C aged and room-temperature deformed specimens gives $\sim$50% as the maximum contribution of the precipitated γ' phase to the total proof stress, whereas carbide precipitated grain-boundaries contribute by only 7%.

(2) In the age-hardening process, the proof stress in general begins to rise much faster than the strain-hardening coefficient. The rise of the proof stress corresponds to the hardening due to shearing of γ' particles by dislocations. Variation with aging time of internal friction of the aged and lightly deformed specimen is similar to that of the strain-hardening coefficient. The critical aging time for the rise of the strain-hardening coefficient and the internal friction corresponds to the development of the microstructure involving dislocation loops and tangles as a result of the interaction of dislocations with coarse γ' particles.

(3) Transmission microstructure observations of the fully age-hardened specimen reveal two types of γ'-dislocation interaction, i.e. shearing and by-passing of γ' particles by dislocations. The existance of both types is due to the effect of the size distribution of γ' particles. Shearing of γ' was also observed in the overaged specimen.

(4) The theoretical expression given by Brown et al. for hardening due to shearing of the γ' particle by paired dislocations gives 20 kg/mm^2 as precipitation hardening for 800°C x 1 hr age-hardened specimen. Addition to this of the contribution of coherency strain-field effect of 5 kg/mm^2 gives the value which is considerably smaller than the measured value of 35 kg/mm^2.

(5) Two types of serrated yieldings appear on the stress-strain curves for the elevated-temperature deformation of Udimet 520. One type is irregular and wavy in form and appears in the range of 200 to 350°C, whereas another type takes place only with the age-hardened specimens; the latter is characterised by the repeated yield drop of comparatively large amplitude and appears in the range of 350 and 600°C. The activation energy for the lower temperature serration is 0.85 eV. and the analysis of the factors affecting the serration suggests that it corresponds to dragging of dislocations by the atmosphere of Al and Ti atoms. The higher temperature serration takes place with the initiation of the plastic deformation; it corresponds to the repeated shearing of γ' particles by dislocations.

(6) Dependence of the proof stress of the age-hardened specimen on deformation temperature is affected by the volume fraction of γ' phase; the specimen fully age-hardened at 800°C contains γ' phase by 26% in volume and the proof stress decreases only slightly with temperature up to 800°C. The comparatively large elevated-temperature proof stress is attributed to the effect of shearing of γ' particles by dislocations and the temperature dependence of γ' strength.

(7) γ'-dislocation interactions in the age-hardened specimen deformed at temperatures below 700 °C involve both shearing and by-passing of γ' particles depending on the size of the γ' particles concerned. On the other hand, when the deformation temperature exceeds 700°C, double cross slip of dislocations around γ' particles takes place and this develops the microstructure which involves dislocation tangles. Deformation of

the specimen at temperatures above 800°C develops the microstructure which involves predominantly tangled dislocations; the rapid drop of the proof stress with temperature is caused by rapid disappearance of γ' shearing dislocations since dislocations invariably cross slip γ' particles.

References

(1) G. P. Sabol and R. Stickler, Phys. Stat. Sol., 35, 11, (1969).
(2) H. Gleiter and H. Hornbogen, Mater. Sci. Eng., 2, 285, (1967/68).
(3) R. F. Decker, Steel Strengthening Mechanisms Symposium, Zurich, (1969).
(4) L. M. Brown and R. K. Ham, Strengthening Methods in Crystals, Elsevier, (1971).
(5) S. M. Copley and B. H. Kear, Trans. Met. Soc. AIME., 239, 984, (1967).
(6) N. J. Petch, J. Iron Steel Inst., 174, 25, (1953).
(7) B. J. Brindley and P. J. Worthington, Met. Rev., 145, 101, (1970).
(8) L. K. Singhal and J. W. Martin, Acta Met., 16, 947, (1968).
(9) B. A. Parker, J. Aust. Inst. Met., 14, 321, (1969).
(10) V. Gerold and H. Haberkorn, Phys. Stat. Sol., 16, 675, (1966).
(11) J. Friedel, Dislocations, Pergamon Press, (1964).
(12) R. A. Swalin and A. Martin, Trans. Met. Soc. AIME., 206, 567, (1956).
(13) O. H. Kriege and J. M. Baris, Trans. ASM., 62, 195, (1969).
(14) P. Beardmore, R. G. Davies and T. L. Johnston, Trans. Met. Soc. AIME., 245, 1537, (1969).

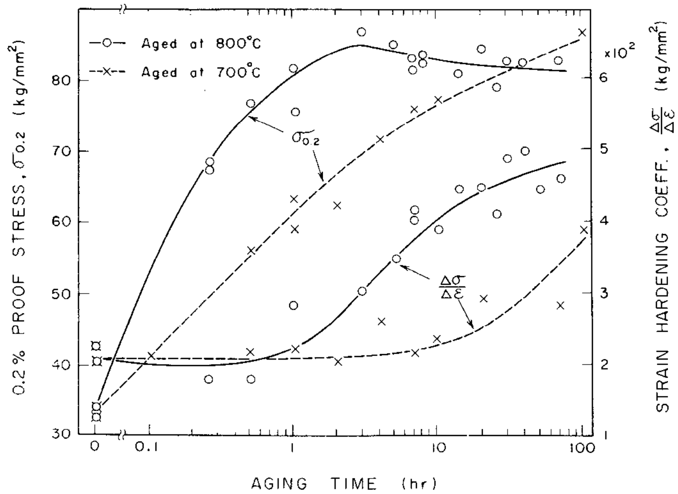


Fig. 1: Variation with aging time of the proof stress and the strain-hardening coefficient of alloy A.

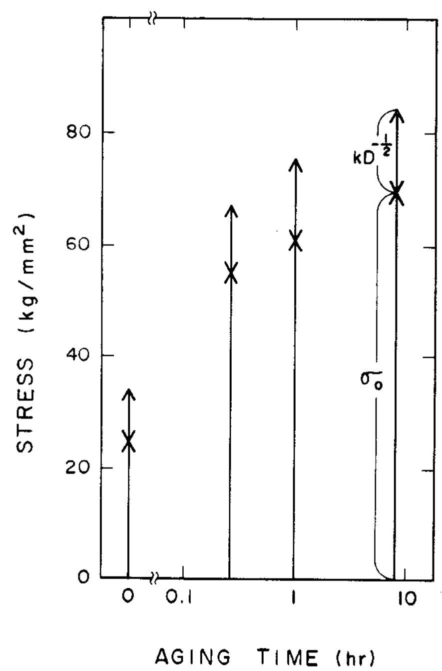


Fig. 2: Variation with aging time of σ_0 and $kD^{-\frac{1}{2}}$ for alloy A obtained from Hall-Petch plots of the proof stress.

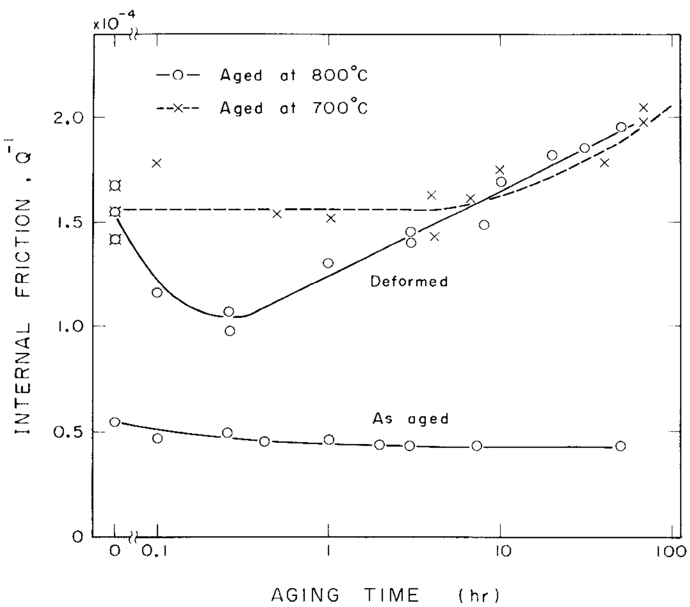
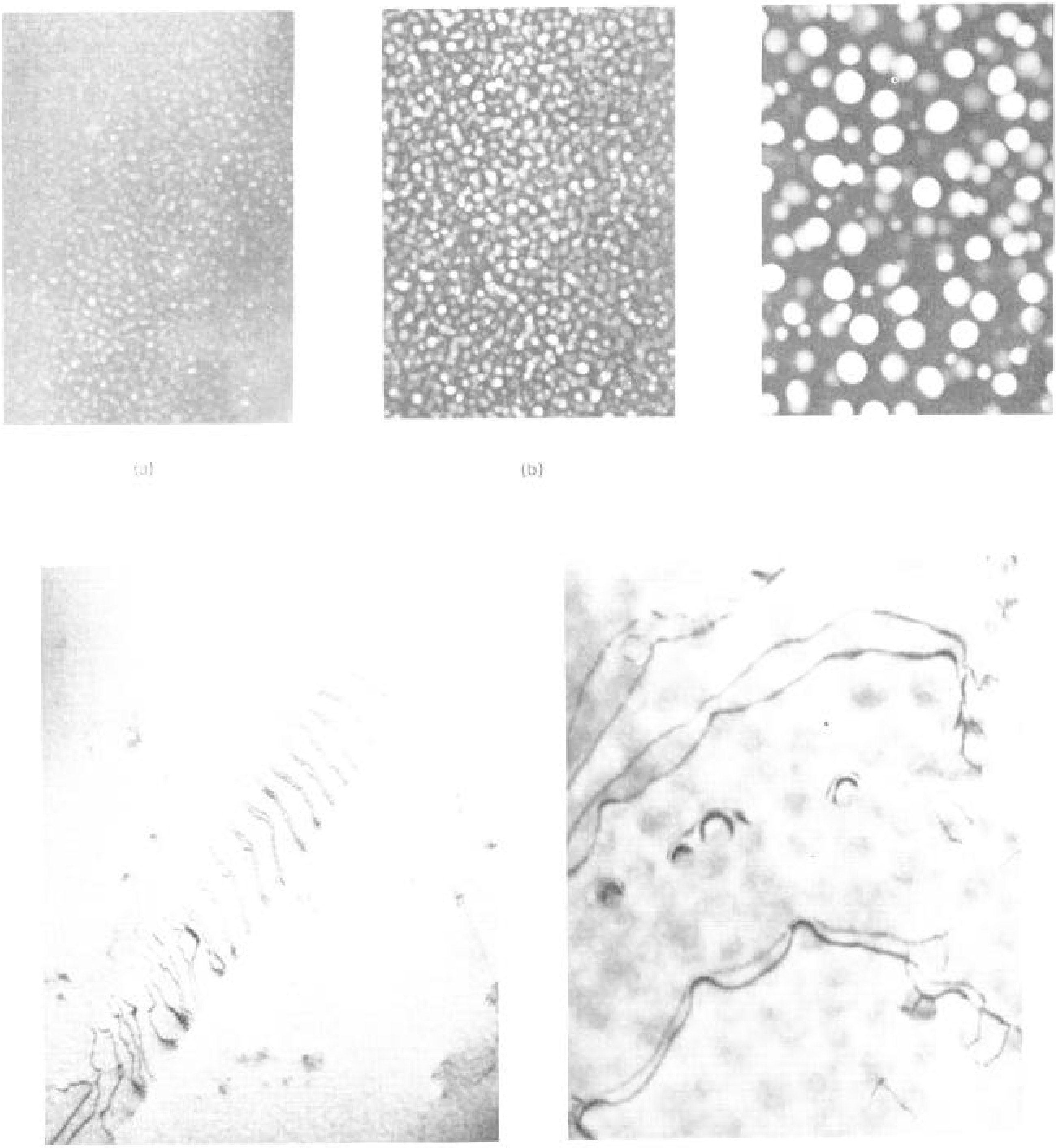


Fig. 3: **Variation with aging time of internal friction of the as aged specimen and the aged and deformed specimens.**



(a) (b) (d) (e)

Fig. 4: Above: Growth of γ' particles of alloy A as a function of aging time at 800°C (dark field images). The aging time is as follows. (a) 0.5 hrs, (b) 3 hrs, and (c) 50 hrs. (x 75,000)
Below: Variation with aging time at 800°C of the arrangement of dislocations in lightly deformed alloy A specimens. The aging time is as follows. (d) 0.5 hrs (x 30,000) and (e) 50 hrs (x 75,000).

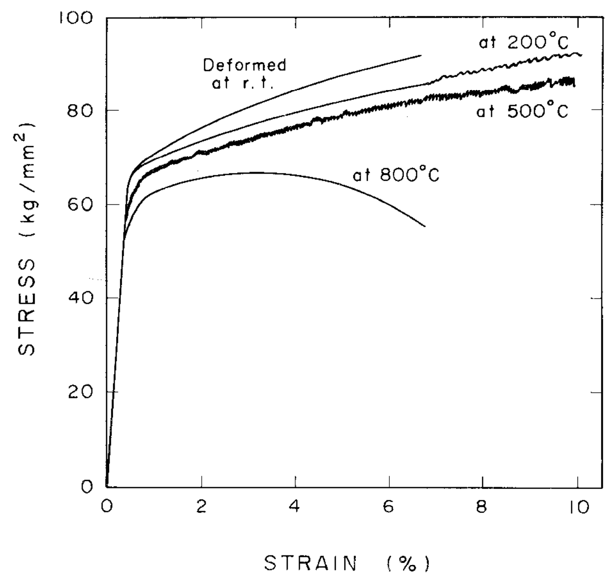


Fig. 5: **Tensile stress-strain curves of alloy B, preliminarily aged for 3 hours at 800°C, for various deformation temperatures.**

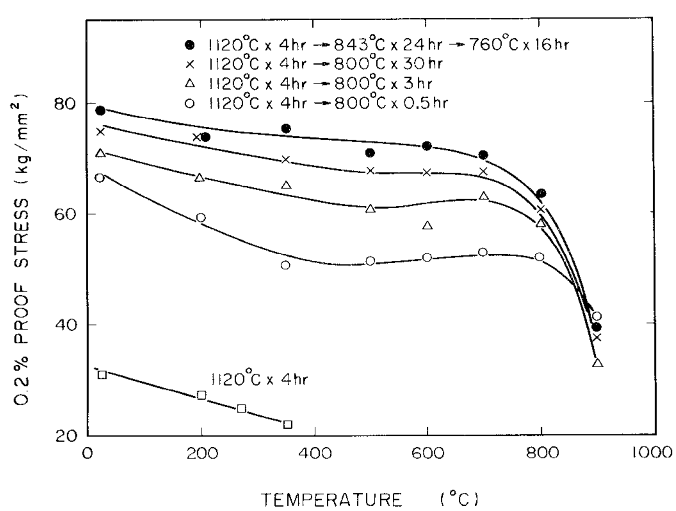
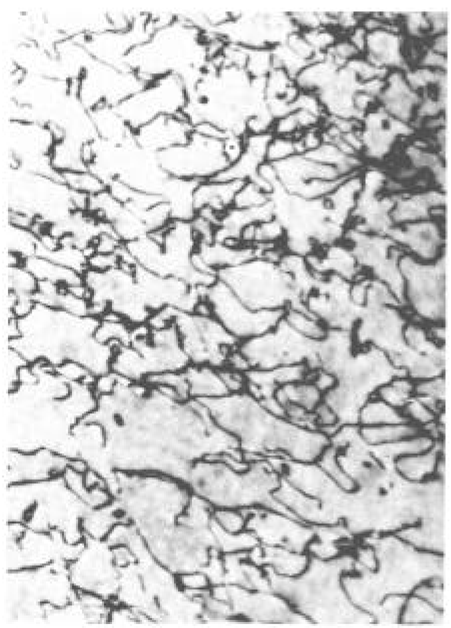


Fig. 6: **Variation with temperature of the proof stress for alloy B specimens preliminarily aged at 800°C for various lengths of time and for the specimen subjected to the two-steps aging treatment.**

(a) Deformed at 500°C. (x 21,000)



(b) Deformed at 900°C. (x 45,000)

Fig. 7: Effect of deformation temperature on the dislocation configuration of alloy B preliminarily aged for 0.5 hours at 800°C.