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Instantaneous plastic strain associated with stress changes during the steady state creep of Al and Al–4.20 at% Mg alloy

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The instantaneous strains resulting from stress changes during steady state creep of polycrystalline aluminium and an Al–4.2 at % Mg alloy in the temperature range 100 to 300° C, have been determined. Instantaneous plastic strains were found in both materials for stress increments and decrements. For polycrystalline aluminium the instantaneous plastic strain on a stress increment, $\Delta\epsilon(+)$, was considerably larger than the instananeous strain on a stress increment, $\Delta\epsilon(-)$, whereas for Al–4.2 at % Mg $\Delta\epsilon(+)$ was approximately equal to $\Delta\epsilon(-)$. Work hardening rates determined from $\Delta\epsilon(+)$ and $\Delta\epsilon(-)$ for polycrystalline aluminium vary from about one-tenth to one-half of Young's modulus and depend strongly on temperature and stress. The need to improve existing creep theories to include both climb (recovery) and glide components is suggested.

1. Introduction

Behaviour upon and after sudden stress changes during steady-rate creep at high temperatures is thought to depend on the type of glide motion of dislocations [1–3]. Oikawa and Sugawara [4] suggest that when the stress is increased by a small amount, plastic strain occurs instantaneously in pure metals in which the glide motion of dislocations approximates free-flight and the creep is believed to be controlled by a recovery process. The work hardening rate, h, can be evaluated from the instantaneous plastic strain associated with the sudden increment in stress. On the other hand, Oikawa and Sugawara [4] consider that for materials whose creep is controlled by the viscous glide of dislocations, instantaneous plastic strain does not occur when the applied stress is increased suddenly. These phenomena are schematically shown in Fig. 1. Note that for materials exhibiting free flight glide it is proposed that the total instantaneous elongation on a stress increment, $\Delta l(+)$, has two components a plastic elongation Δl_p and an elastic elongation, Δl_e. The elastic elongation is considered to be equal to the instantaneous contraction experienced on an equivalent stress reduction, Δl^-. In the case of the material exhibiting viscous glide, $\Delta l^+ = \Delta l^-$, which leaves one with the implicit assumption that there is no plastic component to the instantaneous strain on either a stress increment or on a stress reduction.

Oikawa and Sugawara in their original paper [4] gave some limited data on Al and Al–5.5 at % Mg which appeared to fit these general concepts and behaviour. More detailed data have recently been published by Oikawa *et al.* [5] for polycrystalline aluminium which is in agreement with the behaviour illustrated in Fig. 1a, i.e. elastic instantaneous strain on a stress reduction and elastic plus plastic instantaneous strain on a stress increment. However Langdon and Yavari [6] in their analysis of the original data of Oikawa and Sugawara [4] indicate that although for the Al–5.5 at % Mg alloy the instantaneous strains on a stress increase and the corresponding stress decrease are indeed approximately equal, they are both greater than the strain expected from purely elastic behaviour

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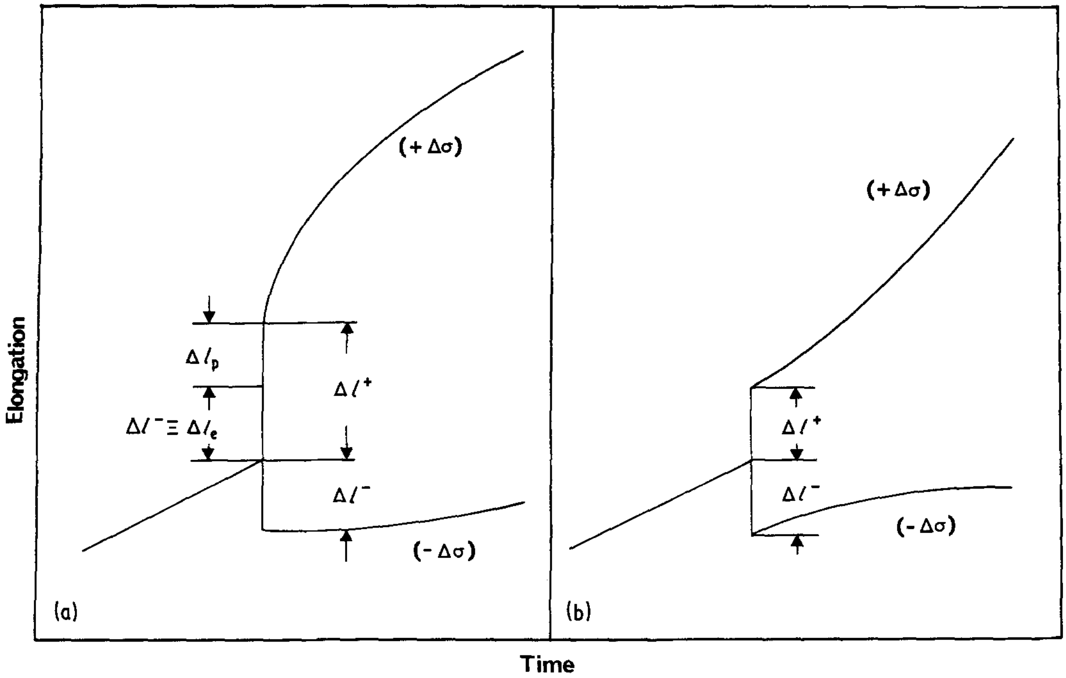

Figure 1 Schematic drawings of the instantaneous elongation and contraction upon small stress changes in cases of (a) Oikawa and Sugawara free-flight glide, and (b) viscous glide [4].

(total instantaneous strain being approximately 3.5 times that expected from elastic behaviour). Langdon and Yavari also showed that Oikawa and Sugawara's data for aluminium at 200° C and an initial stress of 15 MPa in fact indicated an instantaneous strain greater than elastic for both a stress increase and a stress decrease. Langdon and Yavari [6] also presented their own data for Al–5% Zn (behaves like Al with free flight glide) and Al–5 % Mg (viscous glide) to show that the instantaneous strain on a stress decrease was greater than that to be expected from purely elastic behaviour providing the stress reduction was greater than about 25–30 % of the initial stress level. Somewhat similar behaviour was found for super-purity aluminium by Parker and Wilshire [7] where the instantaneous contraction was entirely elastic for stress decrements up to about 50%. If one is trying to determine the work hardening rate, h using the relationship:

$$h = \Delta\sigma/(\Delta l_{\mathrm{pl}}/l) \qquad (1)$$

where $\Delta\sigma$ is stress change, Δl_{pl} is plastic elongation, l is original length and $\Delta l_{\mathrm{pl}} = \Delta l^{+} - \Delta l^{-}$, and if Δl^{-} is considered to be entirely elastic then large errors could arise when Δl^{-} is not entirely elastic. Given the disagreement between the different researchers particularly with regard to the existence of an instantaneous plastic strain on a stress reduction it was considered useful to perform further tests. Pure aluminium and an Al–4.20 at % Mg alloy were chosen for study as being representative of materials exhibiting free-flight glide and viscous glide in steady state creep respectively. Tests were conducted over a range of temperatures (100–300° C) in order to determine whether the behaviour was temperature dependent.

2. Experimental details

The aluminium specimens were of high purity (99.999 %) and were in the form of 2.95 mm diameter wire. The specimens were annealed for 1/2 h at 400° C and furnace cooled in order to remove any residual cold-work. The Al–4.20 at % Mg alloy was vacuum cast using high purity metals. The ingot was swaged and drawn to 2.95 mm diameter wire with appropriate intermediate annealing treatments. The alloy wire specimens were also annealed for 1/2 h at 400° C but then quenched into water to obtain a complete α-phase solid solution prior to creep testing.

A uniaxial constant stress creep machine with high sensitivity was used. The experimental set-up is described in more detail by Smith [8]. All tests were performed in air with a temperature gradient along the 76 mm gauge length of $<1^\circ$ C. Test temperatures of 100, 200 and 300° C were used for the aluminium specimens and 200, 250 and 300° C for the Al–4.20 at % Mg specimens. The specimens were loaded and allowed to reach steady state creep. Steady-state creep rates down to 10^{-8} $\sec^{-1}$ could be measured with confidence and reproductibility. Once steady state creep was obtained various levels of stress increment or decrement were applied and the instantaneous strain measured. Between each stress change the stress was brought back to the initial level and the specimen allowed to approach steady state creep. Three or four tests were conducted for each level of stress increment/decrement. The instantaneous strains were compared to those calculated assuming purely elastic behaviour and using the published values for the elastic moduli of aluminium [9].

3. Experimental results

3. 1. Creep behaviour

The creep curves for aluminium showed a typical normal primary stage with steady state setting in at a true strain of about 0.2. The limited number of tests for aluminium gave a stress exponent of 4 to 4.5 which is typical for aluminium and indicates control by free-flight glide [5]. The Al 4.2 at % Mg alloy showed an inverted primary stage which is typical of Al–Mg alloys containing more than about 3 % Mg [10]. Steady state creep also set in the alloy at a true strain of about 0.2. The stress exponent of steady state creep measured for the Al–4.2 at % Mg alloy at 300° C was 3.68 when calculated using the applied stress, σ_e, and 3.06 when calculated using the effective stress σ_e [11]. A stress exponent of about 3 would indicate control of creep by a viscous glide process [6].

3.2. Instantaneous strain

3.2.1. Aluminium

The instantaneous (total) strains measured at the three temperatures (100, 200 and 300° C) and two different initial stress levels are summarized in Fig. 2. The strain on an increase in load was always larger than that on the corresponding decrease in load. Stress increments were also the most sensitive to the loading method and showed considerable scatter from test to test. The instantaneous elongation increases rapidly with the magnitude of stress increment whereas the smaller contraction is almost linearly proportional to the magnitude of the stress decrement. There were some indications that the instantaneous strain on a stress decrease for small stress changes was approximately equivalent to that expected from elastic behaviour. However, for stress decreases greater than about 10 % (expressed as percentage of initial stress) the instantaneous contraction was always greater than elastic. These observations are in general agreement with those of Oikawa *et al.* [5] except for the very important observation that the instantaneous contraction is greater than that expected from elastic behaviour.

3.2.2. Al–4.2 at % Mg alloy

The instantaneous (total) strains measured at the three temperatures (200, 250 and 300° C) for two different initial stress levels are summarized in Fig. 3. The instantaneous elongation and the instantaneous contraction for equal stress changes are almost the same with the instantaneous elongation being slightly higher. There is an approximately linear relationship between the instantaneous contraction/elongation and the stress change. In general both the instantaneous elongation and contraction are larger than that expected from elastic behaviour. These results agree very well with those of Oikawa and Sugawara for Al–5 % Mg tested at 300° C [4] but differ somewhat from the results of Langdon and Yavari [6] for Al–5 % Mg tested at 400° C where the differences between instantaneous elongation and instantaneous contraction were greater and stress changes greater than about 25% were required for the instantaneous contraction to exceed that expected from elastic behaviour.

4. Discussion and analysis of results

In their analysis of the instantaneous strains in aluminium Oikawa *et al.* [5] calculate an apparent work hardening rate, H, using the expression

$$H = \Delta\sigma/\Delta\epsilon_p \qquad (2)$$

They estimated the plastic strain $\Delta\epsilon_p$ and the difference between the total strain increase $\Delta\epsilon(+)$ on a stress increase and the total strain decrease $\Delta\epsilon(-)$ on a stress decrease of the same amount. $\Delta\epsilon(-)$ was in agreement with the value calculated from Young's modulus of the specimen and the machine stiffness. However in the present work on aluminium, and also in Langdon and Yavari's work on Al–5 % Zn [6] which also exhibits free flight glide, $\Delta\epsilon(-)$ is generally greater than elastic

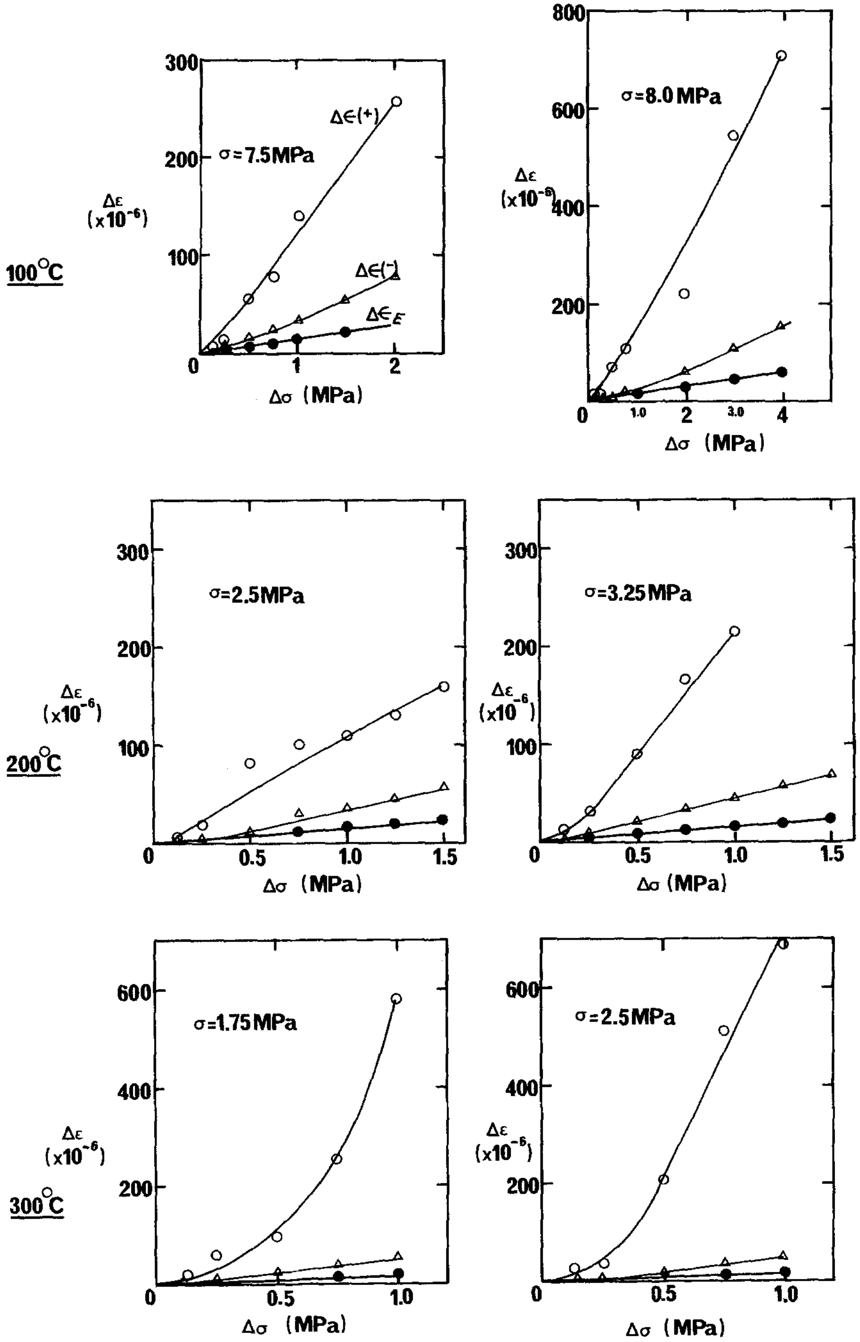

Figure 2 Instantaneous strains in polycrystalline aluminium on stress change during steady state creep. $\Delta\epsilon(+)$[O] is instantaneous strain on stress increment $+\Delta\sigma$; $\Delta\epsilon(-)$[Δ] on stress decrement $-\Delta\sigma$; and $\Delta\epsilon_E$[●] is calculated using Young's modulus.

contraction. Using the rationale of Oikawa *et al.* [5] this would indicate some permanent strain occurs on a decrease in load. There is thus a degree of uncertainty in calculating H using $\Delta\epsilon_p = \Delta\epsilon(+) - \Delta\epsilon(-)$. It would seem more correct to calculate $\Delta\epsilon_p$ using the strain calculated from Young's modulus at the relevant temperature, $\Delta\epsilon_E$ (where E indicates elastic). Therefore $\Delta\epsilon_p = \Delta\epsilon(+) - \Delta\epsilon_E$. For the purpose of the discussion of the results we will designate H calculated from $\Delta\epsilon_p = \Delta\epsilon(+) - \Delta\epsilon(-)$ as H_1 and H calculated from $\Delta\epsilon_p = \Delta\epsilon(+) - \Delta\epsilon_E$ as H_2.

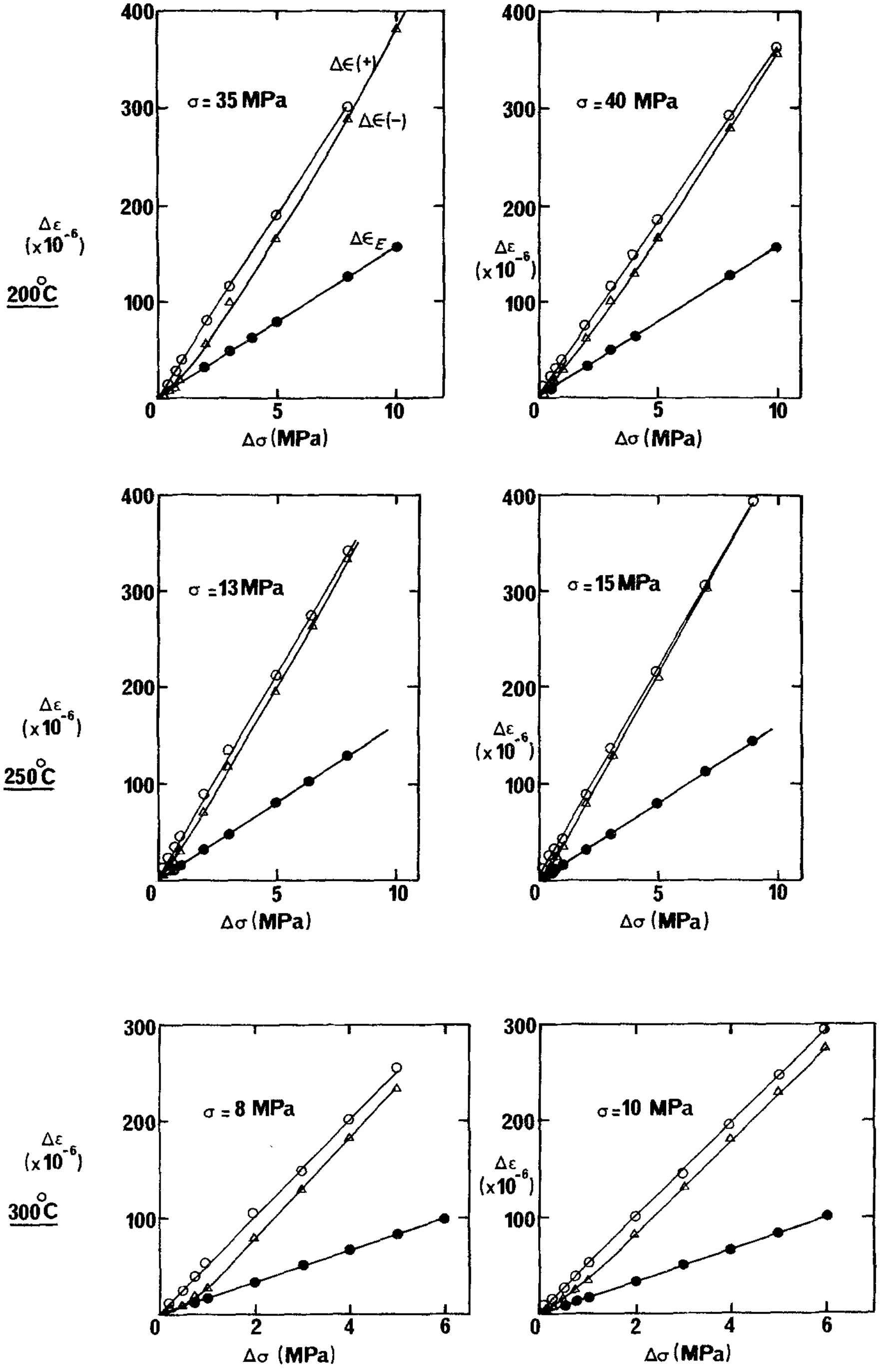

Figure 3 Instantaneous strains in Al–4.2 at% Mg on stress change during steady state creep. $\Delta\epsilon(+)$[O] is instantaneous strain on stress increment $+\Delta\sigma$; $\Delta\epsilon(-)$[Δ] on stress decrement $-\Delta\sigma$; and $\Delta\epsilon_E$[•] is calculated using Young's modulus.

A plot illustrating the dependence of H (in this case H_2) on the stress changes $\Delta\sigma$ is given in Fig. 4 for an initial stress of 3.25 MPa and a temperature of 200° C. The plot is made of $\log H$ against $\Delta\sigma$ since Oikawa *et al.* [5] found that $\log H$ increased linearly as $\Delta\sigma$ decreased. Allowing for the scatter in experimental data, such a dependence is also found for the present results. The true work-hardening rate h, which is not affected by recovery can be estimated as [5]

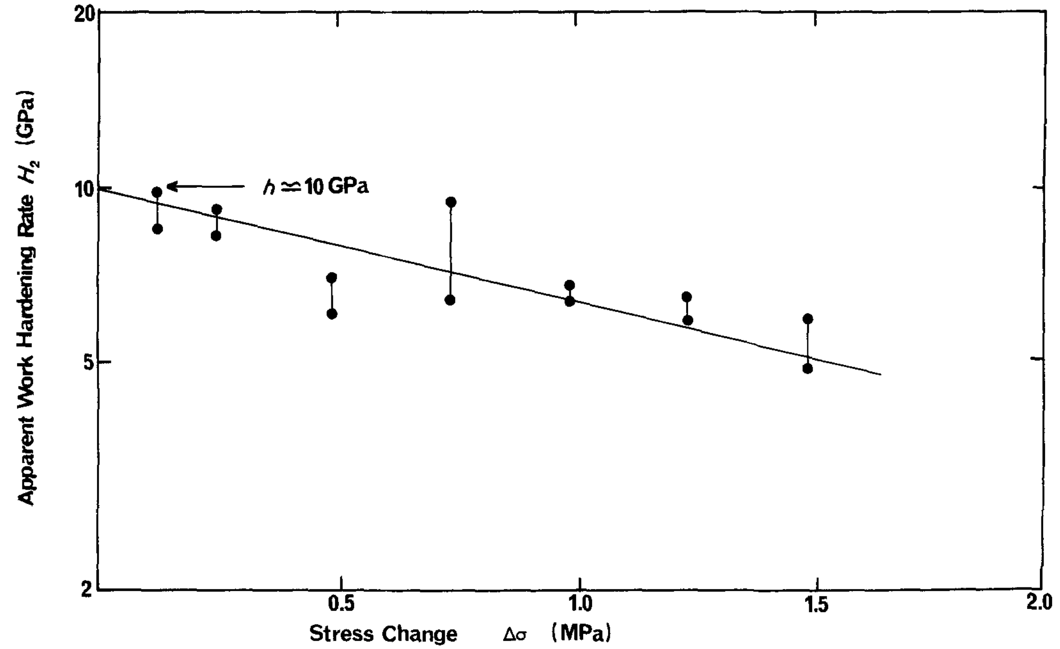

Figure 4 Example of the $\Delta\sigma$ dependence of the apparent work hardening rate H_2 determined from the instantaneous plastic strain that occurs on a sudden stress increase for an intital stress of 3.25 MPa at 200° C.

$$h = \lim_{\Delta\sigma \to 0} H = \lim_{\Delta\sigma \to 0} \left(\frac{\Delta\sigma}{\Delta\epsilon_p} \right)$$

Thus to find h we can extrapolate the $\log H$ against $\Delta\sigma$ plot back to $\Delta\sigma = 0$ as shown in Fig. 4. Such an extrapolation was performed for all initial stresses and temperatures using both H_1 and H_2 and the results are summarized in Table I. The limited number of results for H are replotted in Fig. 5 to illustrate the effect of initial stress and temperature. In agreement with Oikawa *et al.* [5] the value of h decreases with increasing σ, the magnitude of decrease depending on temperature, with lower temperatures producing a larger decrease in h with increasing σ. Assuming h can be represented by the equation, [5] where Q_h is the activation energy

$$h = h_0 \, \sigma^{n_h} \exp -(Q_h/RT) \qquad (4)$$

a stress exponent n_h can be calculated. For the present experiments n_h has values of -0.69, -2.37 and -6.38 for temperatures of 300° C, 200° C and 100° C respectively. These values for n_h plus the data of Oikawa *et al.* [5] and other results determined by Horita and Yoshinaga [12] using the strain rate change method during conventional tensile testing are plotted in Fig. 6 as a function of the testing temperature. The present results at 300° C agree well with the previous Japanese data. It is particularly interesting that our results suggest that the stress exponent of hardening, n_h,

TABLE I Work hardening rates of polycrystalline aluminium obtained from extrapolating apparent work hardening rates to the limit where the stress change $\Delta\sigma = 0$

Temperature (° C)	Initial stress (MPa)	h determined from H_1 (GPa)	h from H_2 (GPa)	Average h (GPa)	$\frac{h_{av}}{E}$
100	7.5	39	32	35.5	0.53
	8.0	24	25	24.5	0.37
200	2.5	22	21	21.5	0.34
	3.25	10	13	11.5	0.18
300	1.75	8	8	8	0.13
	2.5	6	6.5	6.3	0.10

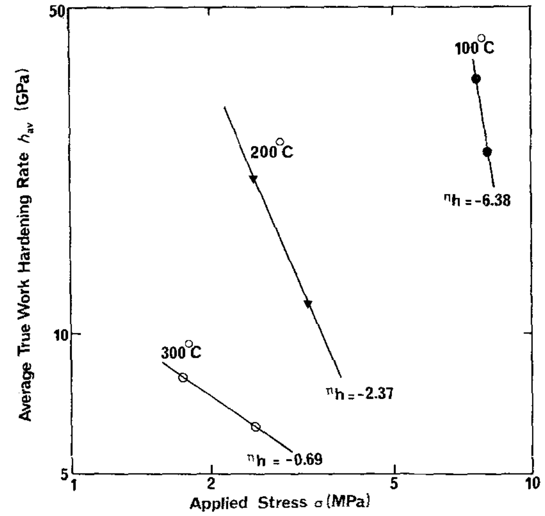

Figure 5 The stress dependence of the average true work hardening rate, h_{av}, during the steady state creep of polycrystalline aluminium.

becomes more highly temperature dependent at temperatures of 200° C or less. This trend was not noted in the previous Japanese work since their lowest testing temperature was 232° C [5].

Our results for the true work-hardening rate, h, are compared in Fig. 7 with previous data for polycrystalline aluminium that has been determined using three different methods. The work hardening rate, h, and the applied stress, σ, have both been normalized by dividing by Young's modulus, E. The data of Oikawa *et al.* [5] were obtained by a similar method to the present results, namely, from stress change tests during steady state creep. The data of Horita and Yoshinaga [12] were obtained using a strain rate change method during conventional tensile testing and those of Sakurai *et al.* [13] by a stress relaxation method during conventional testing. For comparable temperatures lower work hardening rates were found in the present tests than previously. Our

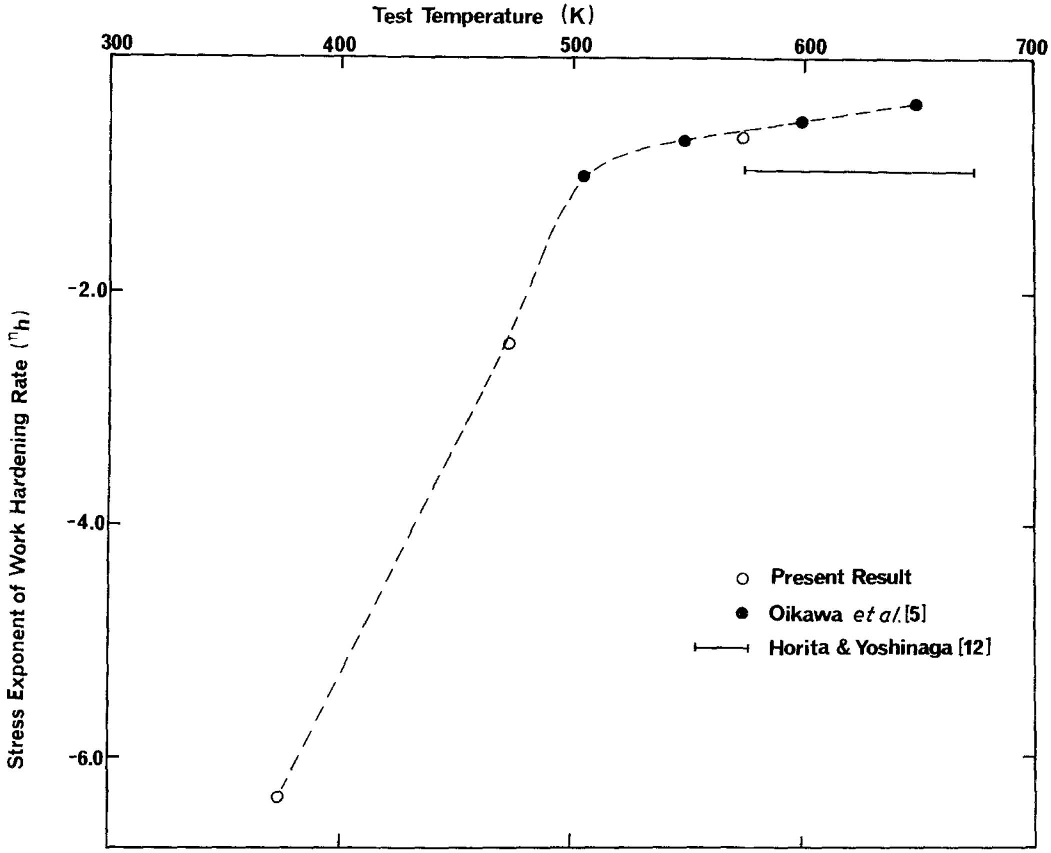

Figure 6 Dependence of the stress exponent of work hardening rate, n_h, on temperature for the steady state creep of polycrystalline aluminium.

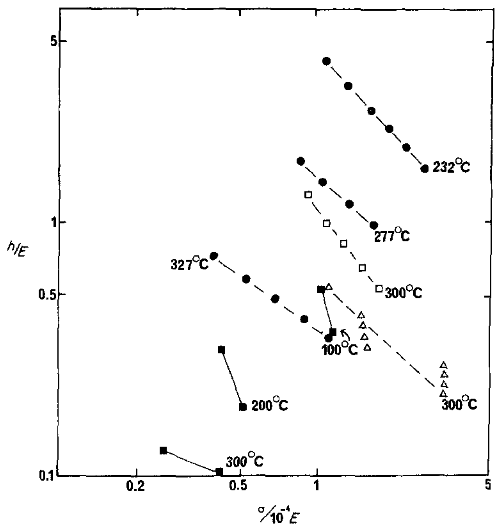

Figure 7 The normalized stress dependence of the normalized true work-hardening rate determined using three methods during the steady deformation of polycrystalline aluminium at high temperatures. ■–■ Present work; ●–● Oikawa *et al.* [5], stress change tests during steady state creep; □–□ Horita and Yoshinaga [12], strain rate change method during conventional tensile testing; △–△ Sakurai *et al.* [13] stress relaxation method during conventional tensile testing.

data at 100° C were comparable to other results obtained at temperatures of about 300° C. Although our results for h are a smaller fraction of E than the Japanese result for polycrystalline aluminium [5, 12, 13] they are more in line with data determined for the high temperature deformation of α-Fe. Ishida and McLean [14] and Davies and Williams [15] determined h for α-Fe at 500–650° C using the plastic strain associated with stress increments during steady state creep and reported that h was about one tenth of E. Watanabe and Karashima [16] also determined h for α-Fe from the slope of the stress–strain curve immediately after creep and found it to be about 0.05 at temperatures from 600 to 750° C.

Thus when looking at our results for polycrystalline aluminium and comparing them principally with the results of Oikawa *et al.* [5] the qualitative agreement is good and some results agree quite well quantitatively. However, there are major differences in that the strain we measured on a stress decrement was greater than elastic whereas only elastic contraction was noted by Oikawa *et al.* [5] and also that the strain hardening rates were significantly lower in our work. In trying to account for the differences it was noted that there were some differences in procedures used in making the measurements of $\Delta\epsilon(+)$ and $\Delta\epsilon(-)$. The procedure of Oikawa *et al.* [5] involved letting the material reach steady state creep, at a stress of σ adding a load $\Delta\sigma$, leaving for 2 sec and then removing the extra load $\Delta\sigma$, and measuring $\Delta\epsilon(-)$. $\Delta\epsilon(+)$ was the instantaneous strain on going from $\sigma \rightarrow \sigma + \Delta\sigma$ and $\Delta\epsilon(-)$ was the instantaneous strain on going from $\sigma + \Delta\sigma \rightarrow \sigma$. Such a method of measuring the instantaneous strains does not fit in with the original concepts for the instantaneous strain measurements as illustrated in Fig. 1 in that although $\Delta\epsilon(+)$ is measured as indicated, i.e. on going from steady state for $\sigma \rightarrow \sigma + \Delta\sigma$, $\Delta\epsilon(-)$ as measured by Oikawa *et al.* [5] is the instantaneous strain going from $\sigma + \Delta\sigma \rightarrow \sigma$ rather than from $\sigma \rightarrow \sigma - \Delta\sigma$. Also when Oikawa *et al.* performed the stress decrement from $\sigma + \Delta\sigma \rightarrow \sigma$ the material was not undergoing steady-state creep at the stress of $\sigma + \Delta\sigma$ since the lapse time of 2 sec is far too short for steady state creep to have been attained before the stress drop. In our tests care was taken to ensure that the steady-state was closely approached at stress σ before either applying a stress increment $(+\Delta\sigma)$ or decrement $(-\Delta\sigma)$. To gauge the importance of the difference in procedures we repeated selected tests measuring $\Delta\sigma(-)$ by both methods i.e. going from $\sigma \rightarrow \sigma - \Delta\sigma$, (our original method) and from $\sigma + \Delta\sigma \rightarrow \sigma$ (Oikawa's method). In general $\Delta\epsilon(-)\sigma + \Delta\sigma \rightarrow \sigma$ was smaller than $\Delta\epsilon(-)\sigma \rightarrow \sigma - \Delta\sigma$. This can be rationalized if one considers the creep rates before the decrement where the specimen would be creeping faster at $\sigma + \Delta\sigma$ than at σ. The faster rate of the forward creep could tend to mask or conceal some of the contraction. However the differences were not that great and $\Delta\epsilon(+)$ was always considerably larger than $\Delta\epsilon(-)$ measured by either method. One must therefore look towards the measurement of $\Delta\epsilon(+)$ to explain any differences in calculated work hardening rates. The measurement is very sensitive to the mode and consistency of loading and judgement as to where the instantaneous strain ends and transient creep begins. Pure aluminium due to its "soft" nature is not the easiest of materials to work with and it is expected that more consistency could be obtained in an alloy was employed. Alloys such as Al–Li or Al–Zn, which also exhibit free-flight glide, may prove more suitable materials for investigation.

In the case of the Al–4.2 at % Mg alloy $\Delta\epsilon(+)$ and $\Delta\epsilon(-)$ are approximately equal, and both are higher than expected from elastic behaviour. This agrees very well with the earlier data of Oikawa and Sugawara [4] for Al–5%Mg. This type of result, i.e. equivalence of $\Delta\epsilon(+)$ and $\Delta\epsilon(-)$ is taken to imply that the glide motion of dislocations is viscous under these experimental conditions [4]. This conclusion derived from the instantaneous plastic strain measurements is in accord with that from other mechanical properties of an Al–5Mg alloy [17] and that from stress exponent measurements of steady-state creep [11].

However, Oikawa *et al.* [5] reason that the presence of instantaneous plastic strain for polycrystalline aluminium on a sudden increase in stress, even when the amount of stress change is small, suggests that there is little resistance to dislocation glide on slip planes, i.e. dislocations glide in a free-flight-like manner. Following on from this they reason that the rate-controlling step or the most time-consuming step is not a process of dislocation glide but a recovery process. They then examine a number of current recovery creep theories [1, 18–22] and find them wanting in many respects. Whilst we would agree with most of the reasoning of Oikawa *et al.* we feel that their analysis itself is somewhat incomplete and incorrect. The presence of an instantaneous plastic strain on a sudden increase in stress is also found for Al–4.2 at %Mg (present work) and for other Al–Mg alloys [4, 6]. Since these materials exhibit viscous glide one can hardly suggest the presence of the instantaneous strain indicates that there is little resistance to dislocation glide on slip planes and that the dislocations glide in a free-flight manner. Furthermore there is an instantaneous strain observed for both Al and Al–4.2 at %Mg on a stress reduction and this must be factored into any creep theory. These observations, together with the fact that the true work hardening rate has an inverse dependence on temperature for aluminium, would suggest that any theory to accurately describe the high-temperature creep behaviour is likely to be extremely complex. We would subscribe that any such creep theory would incorporate both climb (recovery) and glide with the relative importance of each process varying from material to material. We would agree with Oikawa *et al.* [5] that such a theory awaits more high-precision creep tests, such as constant-structure creep tests in the true sense, coupled with the observation of the fine structure developed during creep.

5. Conclusions

Instantaneous plastic strains have been found for both stress increments and stress decrements applied during the steady state creep of polycrystalline aluminium and Al–4.2 at %Mg. For polycrystalline aluminium the instantaneous plastic strain on a stress increment, $\Delta\epsilon(+)$, was considerably larger than the instananeous strain on a stress decrement, $\Delta\epsilon(-)$, whereas for Al–4.2 at %Mg $\Delta\epsilon(+)$ was approximately to $\Delta\epsilon(-)$.

Work-hardening rates determined from $\Delta\epsilon(+)$ and $\Delta\epsilon(-)$ for polycrystalline aluminium at 100 to 300° C vary from about one tenth to one half of Young's modulus and depend strongly on temperature and stress. It is suggested that these results indicate the necessity to improve existing creep theories and that any such creep theory should indicate both climb (recovery) and glide components.

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