

EFFECT OF CHANGING STRAIN RATE ON STRESS-STRAIN BEHAVIOUR DURING HIGH TEMPERATURE DEFORMATION

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Abstract—Plane strain compression tests have been carried out on a ferritic stainless steel at a nominally constant temperature of 917°C and at strain rates of 0.15–5 s⁻¹, and on commercially pure aluminium and an Al–1% Mg alloy at 450°C and strain rates in the range 0.4–12 s⁻¹. Tests have been conducted both at constant strain rate and with strain rate increasing or decreasing from one constant rate to another in a controlled manner after steady state conditions had been established. A wide range of rates of change of strain rate have been studied, with initial and final strain rates differing by up to one and a half orders of magnitude. The ferritic stainless steel and the Al–1% Mg alloy follow a mechanical equation of state under all test conditions of changing strain rate in the sense that the flow stress is dependent only on the instantaneous strain rate and not on the way this strain rate is reached. On the other hand the commercial purity aluminium shows deviations from a mechanical equation of state, which increase systematically with the rate of change of strain rate but are the same magnitude for increasing and decreasing strain rate.

Résumé—Nous avons réalisé des essais de compression en déformation plane sur un acier inoxydable ferritique (pour une température nominale constante de 917°C et des vitesses de déformation variant entre 0,15 et 5 s⁻¹) ainsi que sur de l'aluminium de pureté commerciale et sur un alliage d'aluminium à 1% de magnésium (à 450°C et pour des vitesses de déformation variant entre 0,4 et 12 s⁻¹). Ces essais ont été effectués soit à vitesse de déformation constante, soit à vitesse de déformation croissant ou décroissant d'une vitesse constante à une autre, de manière contrôlée une fois établies les conditions stationnaires. Nous avons étudié un vaste domaine de vitesses de variation de la vitesse de déformation, avec des changements entre vitesses de déformation initiale et finale qui pouvaient aller jusqu'à 1,5 ordre de grandeur. L'acier inoxydable ferritique et l'alliage d'aluminium à 1% de magnésium obéissent à une équation d'état mécanique pour toutes les conditions expérimentales de variation de la vitesse de déformation, dans le sens que la contrainte d'écoulement dépend uniquement de la vitesse de déformation instantanée et non de la façon dont on atteint cette vitesse. Par contre, l'aluminium de pureté commerciale n'obéit pas à une équation d'état mécanique; les écarts observés augmentent systématiquement avec la vitesse de variation de la vitesse de déformation, mais ils sont du même ordre de grandeur pour les vitesses de déformation croissantes ou décroissantes.

Zusammenfassung—Ferritischer rostfreier Stahl wurde bei der nominal konstanten Temperatur von 917°C bei Verformungsraten zwischen 0,15 bis 5 s⁻¹, kommerziell reines Aluminium und die Legierung Aluminium–1% Magnesium bei 450°C und Verformungsraten zwischen 0,4–12 s⁻¹ verformt. Die Versuche wurden mit konstanter Verformungsrate und insofern sich ändernder Verformungsrate durchgeführt, als nach der Einstellung eines stationären Gleichgewichtes von einem konstanten Wert auf einen anderen niedrigeren oder höheren übergegangen wurde. Hierbei wurde eine weite Spanne an Verformungsraten verwendet, mit maximalen Spannen von eineinhalb Größenordnungen. Der ferritische rostfreie Stahl und die AlMg-Legierungen folgen bei allen experimentellen Bedingungen einer mechanischen Zustandsgleichung in dem Sinne, daß die Fließspannung nur von der instantanen Verformungsrate und nicht vom Weg, auf dem diese Rate erreicht wurde, abhängt. Auf der anderen Seite fanden sich bei dem kommerziell reinen Aluminium Abweichungen von der mechanischen Zustandsgleichung; diese nehmen systematisch mit der Änderungsgeschwindigkeit der Verformungsrate zu, sie liegen in derselben Größenordnung bei ansteigender oder abfallender Verformungsrate.

INTRODUCTION

Most laboratory tests for studying flow stress under hot working conditions are performed under constant strain rate and temperature conditions. From such tests it is found that the flow stress (σ) is related to the Zener–Hollomon parameter (Z), which combines the strain rate ($\dot{\epsilon}$) and absolute temperature (T) in a single parameter, as [1]

$$Z = \dot{\epsilon} \exp Q/RT = A (\sinh \alpha \sigma)^n \quad (1)$$

where Q is the activation energy for deformation, R is the universal gas constant, A , α and n are material constants.

A is sensitive to strain before steady state is attained whereas α and n change little with strain and n takes a value of about 5 for most materials. Equation (1) reduces to a power relationship at low stresses and to an exponential relationship at high stresses.

Industrial hot working operations do not take place at constant strain rate and temperature. For

Table 1. Quantometer analyses of the experimental materials (wt%)

	C	Si	Mn	P	S	Cr	Ni	Mo	Cu	Fe	Al	Mg
Ferritic stainless steel	0.058	0.25	0.37	0.034	0.027	20.32	0.31	0.09	—	Rem.	—	—
Commercial aluminium	—	0.06	0.02	—	—	—	—	—	0.02	0.17	Rem.	0.02
Al-1% Mg	—	0.11	0.02	—	—	—	—	—	0.14	0.32	Rem.	0.77

instance, during a rolling pass the strain rate varies from entry to exit as a function of the strain, decreasing rapidly as the maximum strain is approached [2]. Also very steep temperature gradients are created during rolling because the rolls produce a marked chilling effect close to the surface of the material and heating due to the work of deformation is more uniformly distributed through the material [3]. The above changes produce a wide variation in the value of Z during a rolling pass. In the calculation of forces and power requirements for hot working operations, a mean strain rate and mean temperature are usually taken and the flow stress is then derived from relationships such as equation (1). This procedure assumes that the material is following a mechanical equation of state during the change in Z .

Systematic deviations from an equation of state when Z is changing continuously have been reported for pure iron and pure copper [1]. When Z is increasing the flow stress falls below the value expected from an equation of state and the converse is true when Z is decreasing. If Z is changed instantaneously by changing strain rate, Immarrigeon and Jonas [4] found a two stage change in flow stress between steady states in Armco iron and silicon steel. The first stage was instantaneous, but the second required further strain for the flow stress to attain the expected value. However, there was an important difference between the two materials. The pure Armco iron had a relatively large difference between the flow stress at the beginning of the second stage and in steady state, and required a relatively large strain (0.25–0.3) to attain the new steady state value. In the alloyed silicon steel on the other hand, the flow stress immediately after a strain rate increase was only slightly different from the steady state value and attained this value after a small transient strain (0.05–0.08). After a strain rate decrease, no transient was observed in this material. Similar results to those on Armco iron were found on pure aluminium by Young *et al.* [5].

Barraclough and Sellars [6] reported that, after instantaneous changes in strain rate in torsion on a low alloy steel and an austenitic stainless steel, the amount of strain necessary to reach the new steady state stress was usually small and dependent on the magnitude of the strain rate change, on the strain at which it occurs (at strains before steady state is reached) and on the final strain rate. They proposed

a critical value

$$H_{\text{lim}} = \frac{\Delta \log Z}{\Delta \epsilon} \quad (2)$$

for the upper limit of the rates of change of Z that should be allowable in tests where Z varies continuously with strain if transient periods are to be avoided. Values of H_{lim} between 1 and 2 were reported for changes at strains in the steady state range.

The aim of the present work is to analyse in more detail the evolution of flow stress for conditions of continuously changing Zener–Hollomon parameter, and to determine under which conditions, if any, a mechanical equation of state is followed. For this purpose two alloys and a commercially pure metal have been studied using plane strain compression tests. Because changing temperature during the short time intervals available at hot working strain rates leads to temperature gradients in the material, changes in Z have been produced only by controlled changes in strain rate.

EXPERIMENTAL

The three materials employed in the present work are a ferritic stainless steel, a commercial purity aluminium and an Al-1% Mg magnesium alloy of chemical analyses given in Table 1. The three materials were received as 10 mm thick plate, hot rolled for the steel and the Al-1% Mg alloy and cold rolled for the commercial aluminium. Specimens 50 mm wide and 80 mm long for plane-strain compression testing were machined from the plates and then annealed to produce fully recrystallised structures of grain size 760, 100 and 85 μm for the steel, the commercial aluminium and the 1% magnesium alloy, respectively.

Ferritic stainless steel specimens were coated with a glass lubricant (a mixture of Deltaglaze 57† and F1237‡) and allowed to dry for at least 24 h before testing. For the commercial aluminium and Al-1% Mg magnesium alloy, a graphite lubricant (Dag 631†) was brushed on to the specimens just before reheating was to be carried out. These lubricants had previously been found to give a very low coefficient of friction between the tools and the materials during testing [7]. Tests were carried out on a computer-controlled plane-strain compression machine, which has been described previously [7, 8]. Specimens were reheated to the test temperatures for 20 min and tests were performed at nominal tem-

†Acheson Colloids Company, Plymouth.

‡Ramsden and Company, Stoke-on-Trent.

Table 2. Conditions of experimental plane strain compression tests with changing strain rate

	$\dot{\epsilon}_0$ (s ⁻¹)	$\dot{\epsilon}_f$ (s ⁻¹)	H
Ferritic stainless steel	0.15	5.0	6.9, 13.8, 69
	0.5	1.5	4.6, 23
	0.5	5.0	9.2, 23, 46, 115
	1.5	5.0	4.6, 23
	0.5	0.15	-13.8, -20.7
	1.5	0.5	-4.6, -6.9, -9.2, -18.4, -23
	1.5	0.15	-25.3
	5.0	1.5	-4.6, -9.2, -23
	5.0	0.5	-9.2, -13.8, -18.4, -23, -46
	0.4	1.2	21
Commercial aluminium	0.4	4.0	9.2, 32.9, 46
	1.2	4.0	24.2
	1.2	12.0	9.2, 32.9, 46
	4.0	12.0	4.4, 10.1, 21.9
	1.2	0.4	-4.4, -10.1, -21.9
	4.0	1.2	-10.1, -24.2
	4.0	0.4	-4.6
	12.0	4.0	-10.1, -24.2
	12.0	1.2	-4.6
	0.4	4.0	23, 46
Al-1% Mg	4.0	1.2	-12, -24.1

peratures of 917°C for the ferritic stainless steel and 450°C for the commercial aluminium and the aluminium alloy. In practice, the environmental temperature and the tool face temperature were both 917°C for the steel, but at the lower temperature, conduction down the tools led to the tool face temperature being about 40°C below the environmental temperature of 450°C.

Initially, tests at constant strain rates were carried out in the ranges 0.15–5 s⁻¹ for the ferritic stainless steel, 0.4–12 s⁻¹ for the commercial aluminium, and 0.4–4 s⁻¹ for the Al-1% Mg alloy. To study the effects of changing strain rate, tests were performed at a constant strain rate to a strain (ϵ_0) at which the steady state flow stress had been reached, the strain rate was then changed in a controlled manner to a new value, which was then maintained constant to the end of the test. Some tests for the ferritic stainless steel were programmed to change the strain rate linearly

$$\dot{\epsilon} = \dot{\epsilon}_0 \pm K(\epsilon - \epsilon_0) \quad (3)$$

but for most of the tests the strain rate was changed exponentially

$$\dot{\epsilon} = \dot{\epsilon}_0 \exp \pm H(\epsilon - \epsilon_0). \quad (4)$$

The strain rate changes were made over a half, one, and one and a half orders of magnitude with different rates of change measured by K or H . Table 2 summarises the range of conditions studied in the majority of tests, for which the strain rate was changed exponentially with strain. Repeat tests were carried out for many conditions, with specimens being quenched (within 1 s) at different stages during or after the strain rate change to study the evolution of microstructure. The microstructural results will be reported in a separate paper.

During the tests, data on load and displacement were stored in the computer memory at frequencies

appropriate to the strain rate range. After testing, these data were converted to plane strain flow stress and strain using a computer program which incorporates the effect of lateral spread of the specimen during deformation and corrects for the influence of interfacial friction [7]. Strain rate was also computed from the displacements over five sequential data points and the frequency of data readings. Subsequently all stress, strain and strain rate results were converted to the equivalent tensile values using the Levy-von-Mises criteria.

The printouts of the strain rate vs strain obtained directly from the computer attached to the machine were unsatisfactory for the present analysis because the use of a range of five data points in the standard method of calculation still gives random scatter and does not give the true instantaneous value of strain rate when strain rate is changing rapidly with strain. Two alternative procedures were therefore developed to obtain strain rate from the strain vs time data. Method A assumes that equation (4) is exactly followed during the strain rate change. Then, considering time (t) is zero when $\epsilon = \epsilon_0$, integration gives

$$\epsilon = \epsilon_0 - \frac{1}{H} \ln(1 - \dot{\epsilon}_0 H t). \quad (5)$$

The value of H to give the best fit to the experimental strain-time data was computed and then this value of H was used in equation (4) to obtain the instantaneous values of strain rate. This method has the disadvantage that it can only be applied during the strain rate change. Method B was therefore developed in which the strain vs time data were fitted by least squares analysis to a series of polynomial equations. For accurate fitting of the data, it was necessary to divide them into two or three strain ranges and to apply polynomial equations of maximum order five. From the derivatives of these equations the strain rate vs strain curves before, during

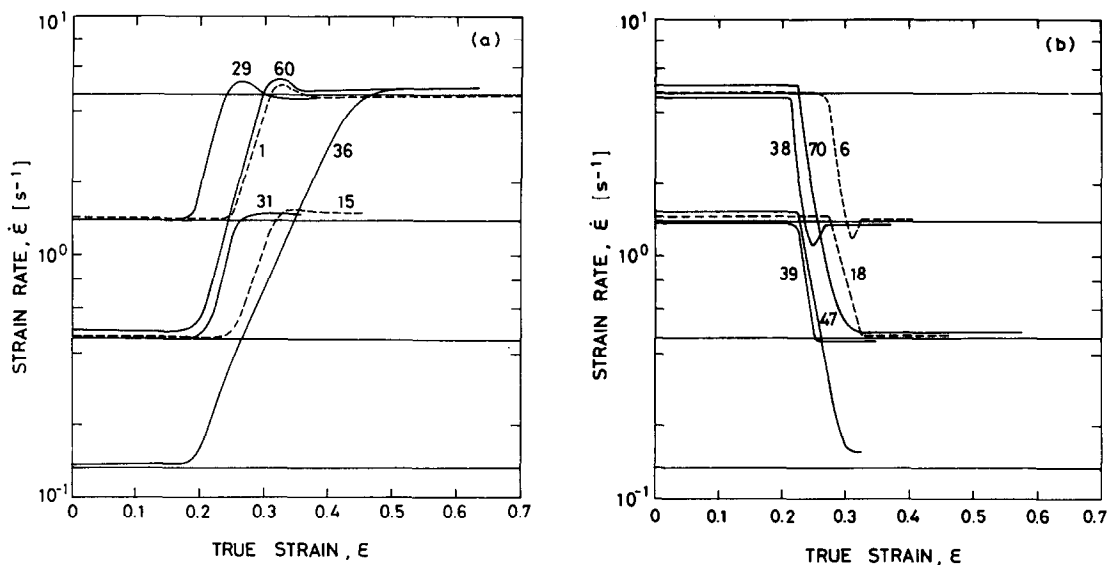


Fig. 1. Types of strain rate change imposed (a) increasing strain rate, (b) decreasing strain rate. Solid lines are for $H = d \ln \dot{\epsilon} / d\epsilon$ constant, broken lines are for $K = d\dot{\epsilon} / d\epsilon$ constant. (Numbers beside the curves indicate the sequential test numbers.)

and after strain rate changes were obtained accurately, even when overshoot and undershoot of strain rate occurred after a change.

RESULTS

The conditions for typical tests with changes in strain rate during deformation are shown in Fig. 1(a) and (b) for increasing and decreasing strain rate respectively. The full lines denote tests in which the logarithm of the strain rate has been changed as a function of strain, and the broken lines tests in which the strain rate has been changed linearly with strain. Although the command signal required the strain rate to remain constant after the change, the response of

the machine led to some overshoot in strain rate for tests with strain rate increasing to 5 s^{-1} and to some undershoot in strain rate for tests with strain rate decreasing to 1.5 s^{-1} . These effects are clearly seen in Fig. 1, as is the fact that at lower strain rates the machine responded closely to the command signals.

Ferritic stainless steel

Figure 2(a) and (b) show the stress-strain curves corresponding to the strain rate changes given in Fig. 1 for ferritic stainless steel tested at 917°C . They also show the stress-strain curves for tests at constant strain rates of 5, 1.5, 0.5 and 0.15 s^{-1} . All these curves have the characteristic that after a short work hardening period the flow stress reaches a steady state

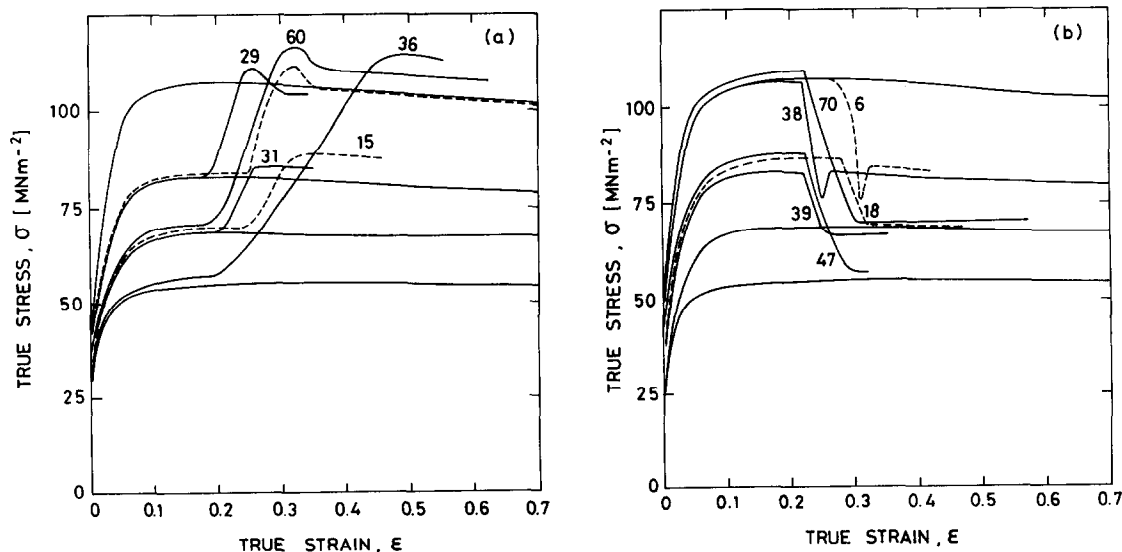


Fig. 2. Stress-strain curves for ferritic stainless steel corresponding to the strain rate changes shown in Fig. 1 in tests carried out at 917°C .

value. At the lower strain rates this remains constant to high strains, but at the higher strain rates, particularly at 5 s^{-1} , there is a fall in stress with increasing strain because of the temperature rise due to the work of deformation. As shown in the figures, all the strain rate changes were made within the strain range of steady flow stress.

For all the tests with changing strain rate, it can be seen that the flow stress closely follows the change in strain rate. This is true even when overshoot or undershoot of the strain rate occurs after the change. It can be seen that there is some variation in the initial flow stress levels from test to test because of small variations in the temperature about the nominal value of 917°C in individual tests. Generally this means that tests with higher values of stress before the strain rate change have higher values of flow stress after the change. It is, however, noticeable that in tests 60 and 36, in which strain rate was increased from a low value to 5 s^{-1} , the flow stress is significantly higher than the steady state level at 5 s^{-1} . As will be discussed later, this is due to the difference in temperature at high strains for the changing and constant strain rate tests. It can be seen qualitatively that, because the deformational heating is roughly proportional to the area under the stress-strain curve, it will be lower for the specimen deformed with changing strain rate than for the specimen deformed at constant strain rate, leading to higher flow stresses.

Figure 3 shows the effect of imposing different rates of change of strain rate ($H = d \ln \dot{\epsilon}/d\epsilon$) on the stress-strain curves for the ferritic stainless steel at 917°C . In this figure it can again be observed that the change in flow stress reproduces the imposed change in strain rate, both with increasing and decreasing strain rate even at the highest imposed rates of change. Again the overshoot in strain rate is followed by the flow stress and the differences in flow stress at the end of the tests with increasing strain rate arise

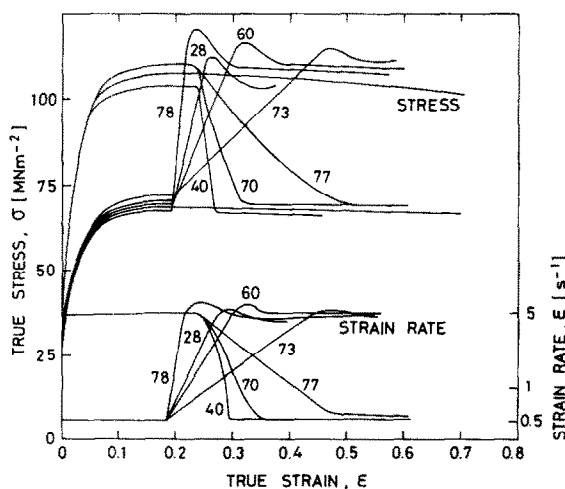


Fig. 3. Effect of imposing different values of $H = d \ln \dot{\epsilon}/d\epsilon$ on the stress-strain curves for ferritic stainless steel at 917°C .

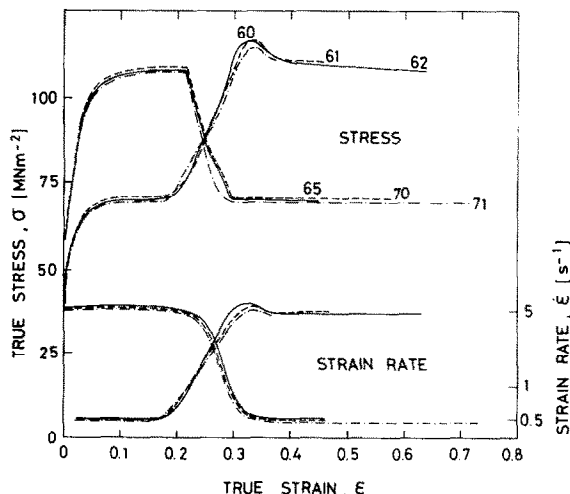


Fig. 4. Reproducibility of stress-strain curves for ferritic stainless steel tested under nominally identical conditions of changing strain rate at 917°C .

from differences in temperature due to deformational heating.

Figure 4 shows the reproducibility of stress-strain curves obtained under nominally identical conditions of changing strain rate but with deformation stopped at different strains for metallographic study. It is clearly apparent that when tests are from a single series where initial temperatures are identical, the results are reproducible and that the small differences in strain rates between the different tests are reflected in the corresponding flow stress curves.

All these results indicate that in the ferritic stainless steel there is a unique relationship between the flow stress and the instantaneous values of strain rate and temperature, which is not dependent on the strain rate history, i.e. whether a constant strain rate or a changing strain rate has been applied. This is shown more clearly in the next section in which the results are analysed in detail.

Only a few tests with changing strain rate were carried out for the Al-1% Mg alloy, but the results showed identical characteristics to those of the ferritic stainless steel.

Commercial aluminium

Figure 5 shows the type of strain rate changes and the corresponding stress-strain curves for commercial aluminium tested at 450°C . At first sight, the flow stress results look similar to those described above, but close inspection of the curves shows that there is a delay in the flow stress reaching the value expected for the changing strain rate. Thus during the strain rate change in tests with increasing strain rate, the actual flow stress is always lower than the value corresponding to tests at the same constant strain rate and the opposite is true for tests with decreasing strain rate, the differences being larger as $H = d \ln \dot{\epsilon}/d\epsilon$ becomes higher. In other words, for the commercial aluminium, the flow stress is not only

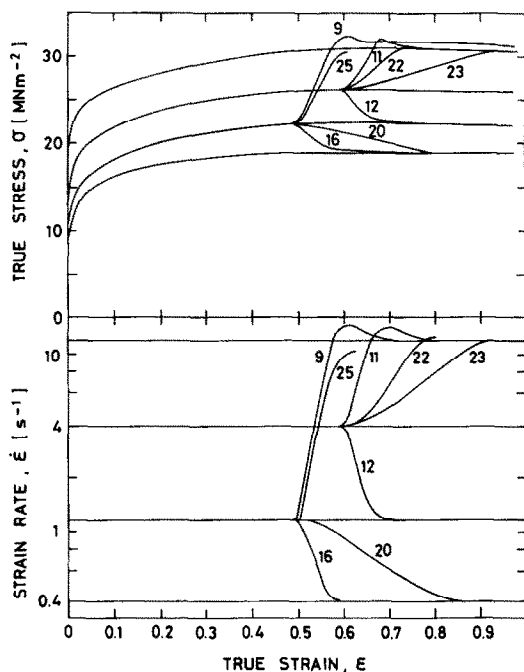


Fig. 5. Types of strain rate changes and resulting stress-strain curves for tests on commercial aluminium carried out at an initial temperature of 450°C.

dependent on the strain rate, but also on the way in which the strain rate has been reached. This behaviour is quite different from that for the ferritic stainless steel, and Al-1% Mg alloy, as will be shown more clearly in the next section.

ANALYSIS OF RESULTS

Constant strain rate

In Fig. 6 the logarithm of the strain rate is plotted against the logarithm of the steady state flow stress. The experimental points for each material fall on straight lines, which means that a power law

$$\dot{\epsilon} = A \sigma^n \quad (6)$$

is followed by these materials in the analysed stress range. To avoid the use of complicated units for A , equation (6) is assumed to be $\dot{\epsilon} = A (\sigma/1)^n$ and then A takes the units of s^{-1} independent of n . Regression analysis gave the values $A = 2.4 \times 10^{-10} s^{-1}$ and $n = 5.1$ for the ferritic steel, $A = 4.2 \times 10^{-10} s^{-1}$ and $n = 7.0$ for the commercial aluminium and $A = 9.4 \times 10^{-14} s^{-1}$ and $n = 7.8$ for the Al-1% Mg alloy. The results were also plotted as the logarithm of strain rate vs the flow stress. For each material the points clearly fell on curves, indicating that equation (6) is a better representation of the data than an exponential relationship between stress and strain rate.

As mentioned in the experimental section, for tests on the commercial aluminium and the aluminium alloy the tool face temperatures were below the environmental furnace temperature of 450°C so that

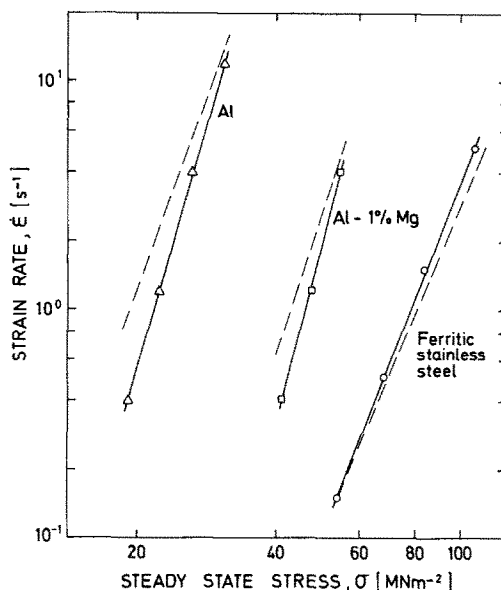


Fig. 6. Relationship between strain rate and steady state stress for the three experimental materials. The broken lines show the relationships expected for true isothermal conditions of deformation.

some cooling of the specimens occurred during testing, particularly at the lower strain rates. The expected temperature changes were computed using a finite difference heat flow model developed by Foster [9]. This showed that at the highest strain rate for commercial aluminium of $12 s^{-1}$ the cooling was nearly compensated for by the deformational heating, leading to a net fall of only 3°C by a strain of 1.0, i.e. essentially isothermal test conditions. However, at lower strain rates the cooling effect became more significant; 13, 18 and 23°C at a strain of 1.0 for strain rates of 4.0, 1.2 and $0.4 s^{-1}$, respectively. Thus, at the lower strain rates particularly, the cooling effect contributes significantly to the extended strain before steady state is achieved at constant strain rate, which is apparent in Fig. 5.

If equation (6) is rewritten in terms of the Zener-Hollomon parameter as

$$Z = \dot{\epsilon} \exp Q/RT = A' \sigma^n \quad (7)$$

where Q is the activation energy for deformation and T is the instantaneous value of mean absolute temperature, then the equivalent strain rate for isothermal deformation conditions can be calculated. Using an activation energy of 156 kJ/mol for aluminium and Al-1% Mg leads to the broken lines in Fig. 6. These show directly the effect of the chilling at a strain of 1.0 and give values of the exponent n' of 5.6 and 6.4 for the aluminium and Al-1% Mg respectively.

In the case of ferritic steel, the tool faces and environmental furnace temperature were the same and there was no chilling effect. At the lower strain rates deformation was essentially isothermal and the same analysis was applied as for aluminium. At the

highest strain rate of 5 s^{-1} the deformational heating could not be neglected and correction was again made by using the Zener-Hollomon parameter with an activation energy of 240 kJ/mol [7]. This led to the broken line with a stress exponent n' of 4.9.

Changing strain rate

Computations were also carried out to determine the influence of changing strain rate on the cooling rate, but over the strain range of interest the effects were found to be negligibly small so the subsequent analysis for aluminium and Al-1% Mg has not been made in terms of Zener-Hollomon parameter but has been made by considering the change in strain rate only. To do this, equation (6) has been applied as an equation of state with the values of A and n derived for each individual test from the flow stress just before and significantly after the change in strain rate. For ferritic stainless steel, the same procedure was adopted except for tests with a final strain rate of 5 s^{-1} , when the deformational heating had to be taken into account.

To look for departures from a mechanical equation of state, equation (6) was applied in three ways. The first two made use of the instantaneous strain rates computed by Method A and Method B, as described earlier. Equation (6) was then applied to obtain the calculated values of instantaneous flow stress for comparison with observed flow stresses. Method B has the advantage that comparison can be made both during and after the strain rate change. However, because of the high values of n in equation (6) it was felt that small departures from a mechanical equation of state could still be masked. An alternative Method C was therefore also applied. For this the observed flow stresses were transformed to instantaneous strain rates using equation (6) and these were then integrated for comparison with the observed total strain as a function of time.

Ferritic stainless steel

Figure 7 gives examples of the results of methods of analysis A and B compared with the observed stress-strain curves for increasing and decreasing strain rate conditions. In all cases when the applied $H = d \ln \dot{\epsilon} / d\epsilon$ was constant, an exponential relationship between the strain rate and the strain was closely fitted by the experimental data and method A gave close agreement between the calculated and observed flow stresses during the strain rate changes, as illustrated in Fig. 7. Method B gave similar good agreement both during and after the strain rate changes, indicating that a mechanical equation of state is followed throughout. This is confirmed by analysing the data using method C, as illustrated in Fig. 8.

Figures 9(a) and (b) compare the experimental flow stress during very steep strain rate changes for increasing and decreasing strain rate with those calculated by method B. In Fig. 9(a) it is observed that the

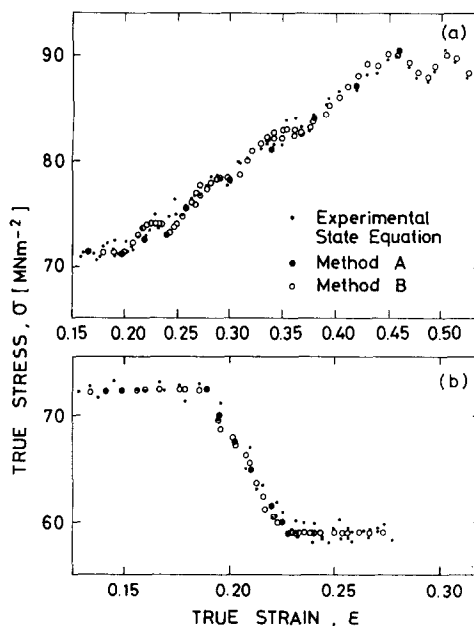


Fig. 7. Comparison of experimental stress-strain data with values calculated using an equation of state applied by methods A and B for ferritic stainless steel tested at 917°C with (a) strain rate increasing from 0.5 to 1.5 s^{-1} , $H = 4.6$ and (b) strain rate decreasing from 0.5 to 0.15 s^{-1} , $H = -20.7$.

agreement between the experimental and calculated points assuming a constant temperature (round symbols) is good at the beginning of the change, but at the end of the change the experimental flow stresses are higher than the calculated ones. Computations using the finite difference program of Foster [9] showed that during and after the strain rate change the deformational heating caused a temperature rise of 8°C . Using the change in mean temperature to obtain instantaneous values of Zener-Hollomon parameter and applying this instead of strain rate in the

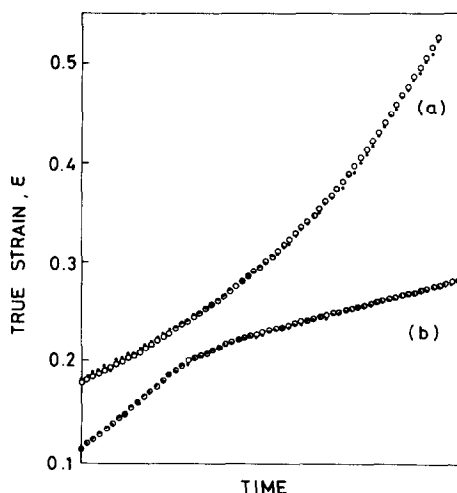


Fig. 8. Comparison of experimental strain-time data with values calculated using an equation of state applied by method C for the test conditions of Fig. 7.

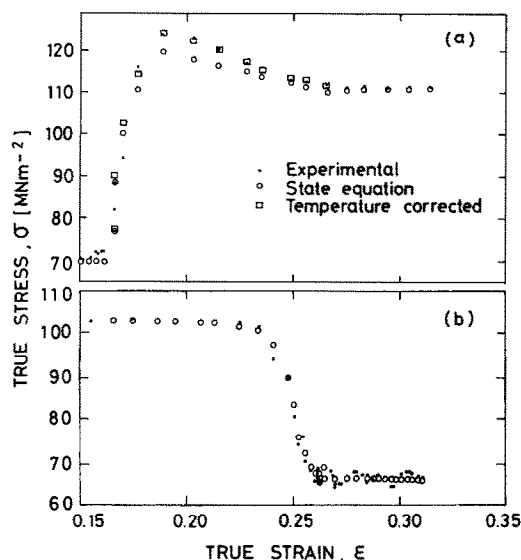


Fig. 9. Comparison of experimental stress-strain curves with curves calculated from strain rate using an equation of state applied by method B for ferritic stainless steel tested at an initial temperature of 917°C with (a) strain rate increasing from 0.5 to 5.0 s⁻¹, $H = 115$ and (b) strain rate decreasing from 5.0 to 0.5 s⁻¹, $H = -46$.

equation of state gave the calculated values shown by the square symbols. These are now in excellent agreement with the experimental points throughout and after the strain rate change. In Fig. 9(b) the temperature change during and after strain rate change was negligible and close agreement is obtained between the points calculated from the strain rate change and the observed flow stresses. Thus a mechanical equation of state between flow stress and Zener-Hollomon parameter is closely followed, a conclusion confirmed by the coincidence of calculated and observed points for each method of analysis for all other test conditions, with the possible exception of one. The exceptional test had a decrease of strain rate from 5 to 0.5 s⁻¹ with $K = d\epsilon/d\dot{\epsilon} = -90$ s⁻¹. In this case the equation of state applied during the change, but after it there was severe undershoot of the strain rate with a short period when it apparently fell to zero. During this period the stress did not fall to zero, indicating a breakdown of the equation of state when the rate of change of strain rate with strain becomes excessively high. It was not possible to assess the value of this limit nor to be certain that the effect was not caused by a limitation in machine response in this case.

Aluminium-1% magnesium

Typical analyses of the results for Al-1% Mg are shown in Fig. 10 for increasing and decreasing strain rates. As for the ferritic stainless steel, it can be seen that a mechanical equation of state is closely followed both during and after the strain rate changes. This was confirmed by analysis of the other test results for this alloy.

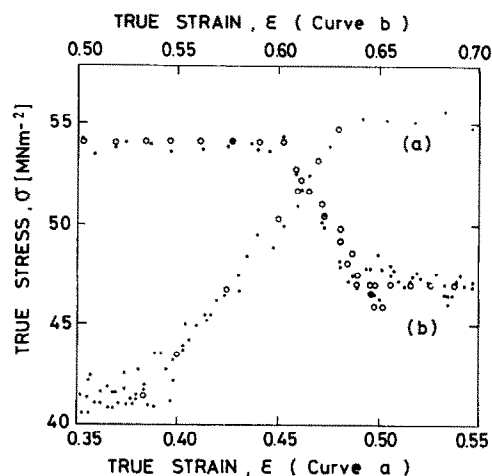


Fig. 10. Comparison of experimental stress-strain data with values calculated using an equation of state for Al-1% Mg magnesium tested at an initial temperature of 450°C with (a) strain rate increasing from 0.4 to 4.0 s⁻¹, $H = 23$, and (b) strain rate decreasing from 4.0 to 1.2 s⁻¹, $H = 24.1$.

Commercial aluminium

Figure 11 compares the experimental stress-strain curves with those calculated by method B for tests with increasing and decreasing strain rate. It is clearly apparent that the experimental flow stresses do not follow an equation of state for either type of test, falling below the calculated values when strain rate is increased and above the calculated values when strain rate is decreased. Figure 12 shows the same data analysed by method C. Again, the departures from an equation of state are clearly apparent. This type of behaviour was observed for all the changing strain rate tests on commercial aluminium.

In order to evaluate the deviation of the experimental flow stress from the value calculated from the strain rate assuming an equation of state during the strain rate change, the experimental points, e.g. in Fig. 11, have been fitted to an equation of the form

$$\ln \sigma_{\text{exp}} = a + b(\epsilon - \epsilon_0) \quad (8)$$

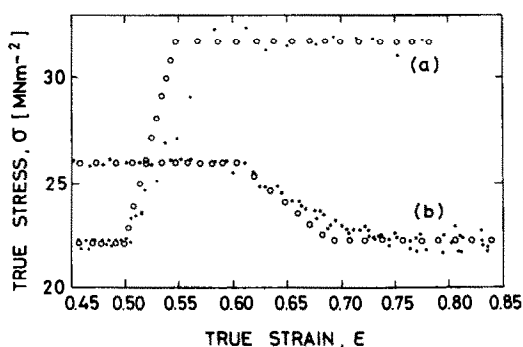


Fig. 11. Comparison of experimental stress-strain data with curves calculated using an equation of state applied by method B for commercial aluminium tested at an initial temperature of 450°C with (a) strain rate increasing from 1.20 to 12.0 s⁻¹, $H = 46$ and (b) strain rate decreasing from 4.0 to 1.20 s⁻¹, $H = -10.1$.

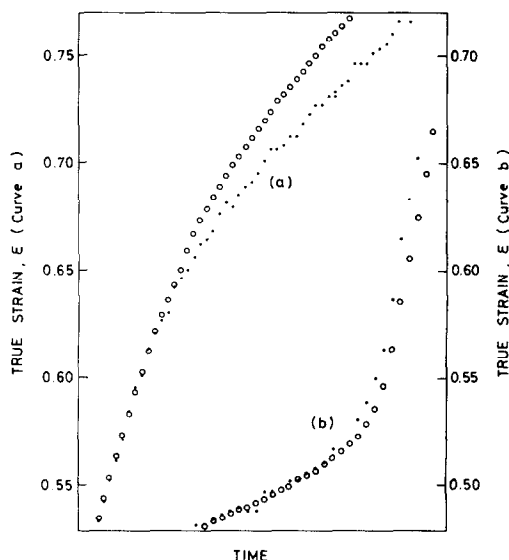


Fig. 12. Comparison of experimental strain-time data with values calculated using an equation of state applied by method C for the test conditions of Fig. 11.

where ϵ_0 is the strain at the start of the strain rate change, and the constants a and b have been determined for each test by regression analysis.

From the experimental strain-time data an exponential equation was fitted, as described for analysis method A, during the strain rate change to obtain the constant H for each test in the relationship

$$\ln \dot{\epsilon} = \ln \dot{\epsilon}_0 + H(\epsilon - \epsilon_0). \quad (9)$$

The values of H obtained were always close to those set in the control program, but were systematically slightly lower for increasing strain rate and slightly higher for decreasing strain rate.

From equations (8) and (9), the constants m and B in the relation:

$$\dot{\epsilon} = B\sigma_{\text{exp}}^m \quad (10)$$

can be obtained as

$$m = H/b \quad (11)$$

and

$$B = \dot{\epsilon}_0 \exp - ma. \quad (12)$$

Equation (10) can now be compared directly with the relationship

$$\dot{\epsilon} = A\sigma_{\text{th}}^n \quad (13)$$

obtained for each test from the values of strain rate and flow stress just before the change and after the new steady state has been obtained. This gives the value of flow stress (σ_{th}) expected if an equation of state were followed during the strain rate change.

Figure 13 shows the ratio, n/m , of the exponents in equations (13) and (10) vs the modulus of the rate of change of strain rate $H = d \ln \dot{\epsilon} / d\epsilon$ for the three experimental materials. A value of $n/m = 1$ means that a mechanical equation of state is followed, as

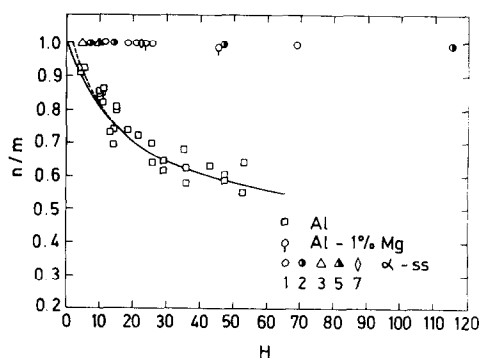


Fig. 13. Dependence of n/m on rate of change of strain rate, H , for the experimental materials. (Numbers beside symbols for ferritic stainless steel indicate the number of tests for a given value of H .)

shown by the data for ferritic stainless steel and aluminium 1% magnesium to the highest values of H applied. In contrast, for commercial aluminium, the ratio is always less than unity and decreases with increasing H , independent of the sign of H and of the initial strain rate or range of the strain rate change, apparently approaching an asymptotic value of about 0.5 for high values of H .

DISCUSSION

The relationship between flow stress and strain rate for constant strain rate tests on ferritic stainless steel, shown in Fig. 6, compares closely with the results previously published for steels of 18–25% Cr [7, 10–12]. The results for commercial aluminium fall in the middle of the range of the results reported by Wong and Jonas [13] for commercial aluminium and high purity aluminium, being very close to the results of Gutierrez [14] on a commercial aluminium, when the present results are corrected for the temperature change during the tests as described earlier. Finally, the results on Al-1% Mg corrected for the temperature change are very close to those reported by Puchi [15] for a similar alloy.

The behaviour of commercial aluminium when strain rate is changed progressively is similar to that observed in the same material when a sudden change in strain rate is made [5] or when in creep tests a sudden change in applied stress is made [16]. In the latter tests a decrease in stress initially gives a much lower value of strain rate than the steady state one corresponding to the new stress, and the converse is true for an increase in stress. This leads to values of $n < m$ in equations equivalent to equations (10) and (13). Similar behaviour has been observed in a broad group of pure metals: Al [17], Ni [17], Cu [18], W [5], Au [19], in some alloys of the group Al-Zn [20] and in NaCl [5].

If the results of the present work shown in Fig. 13 are extrapolated to high values of H , which will be the equivalent situation to a sudden strain rate change, an asymptotic value for n/m of about 0.5 is obtained.

This value is in close agreement with the n/m relationships reported in the literature and summarized by Young *et al.* [5], where values between 0.47 and 0.71 are observed for several materials. Also the value of 0.5 coincides with that recently reported by Blum and Finkel [21] for Al-11 wt% Zn. Thus, the commercial aluminium analysed in the present work follows closely the behaviour reported for other pure metals and class II alloys, in which climb is the main mechanism controlling the deformation.

In the case of ferritic stainless steel and the Al-1% Mg alloy, their behaviour summarized in Fig. 13 does not agree with the normal transient explained above, with $n < m$, nor with the inverse transient reported by Horiuchi and Otsuka [16] for Al-3 at% Mg and Al-6.9 at% Mg, and observed generally for class I alloys, where $n > m$. Although n is higher than m in the results of Horiuchi and Otsuka, the inverse transients observed in their figure are smaller than the normal transients for the high purity aluminium. For a stress change giving an order of magnitude change in strain rate, the strain rate just after the change is 2 times the strain rate for steady state in tests with decreasing stress and 1/2 in tests with increasing stress. Assuming a value of $n = 5$, the corresponding relation in stress would be $2^{1/5} \cong 1.15$. This value is much larger than the limit of detectability of differences in stress in the present work, where a ratio lower than 1.02 can be detected. Thus, clearly the present results on ferritic stainless steel and Al-1% Mg do not agree with the behaviour observed in pure metals and class II alloys (climb controlled) or in class I alloys (glide controlled by viscous drag). Their behaviour falls midway between the two extremes.

A behaviour similar to this was suggested by Immarigeon and Jonas [4] when they analysed the results on an iron-silicon alloy compared with those on Armco iron after a sudden change of two orders of magnitude in strain rate. Although they did not find an instantaneous change of flow stress to the new steady state value in the silicon iron, the instantaneous stress was close to it, and the new equilibration stress was reached after a small transient at the new strain rate; $\epsilon = 0.05-0.08$ compared with a strain of 0.25-0.3 for the Armco iron. In terms of the present continuously changing strain rate tests, the strain interval of 0.05-0.08 suggests that values of H higher than about 60-90 would be necessary to observe a transient in this material. This compares with the experimental values of H up to +115 and -46 applied to the present ferritic stainless steel without the occurrence of transients. However, some evidence for breakdown of the mechanical equation of state at even higher rates of change of strain rate was obtained from a test with $K = d\epsilon/d\dot{\epsilon} = -90 \text{ s}^{-1}$. Such breakdown above some effective upper limit of rate of change of strain rate is to be expected [6] unless infinitely rapid stress relaxation can occur. In Fig. 13 for commercial aluminium, a continuous departure

of the curve of n/m from unity is shown for all values of H . However, with the scatter of results, it is possible that the curve should have a real intercept of, say, $H \sim 2$ below which an equation of state is effectively obeyed. This curve also suggests that once the critical value of H is exceeded for the other materials they may depart rapidly from an equation of state.

The mechanical equation of state is followed by the ferritic steel and aluminium magnesium alloy even though the microstructure does not change instantaneously, i.e. an equation of state for dislocation structure is not followed [22]. This can be explained in terms of a new model [23], developed by taking account of the interaction between dislocation climb and glide in high temperature deformation. For alloys in which the velocity of climb and the velocity of glide are of the same order of magnitude, both mechanisms become so dependent on each other that the extreme models of climb control and glide control cannot be applied individually to them. In this case the new model shows that, within certain limitations, stress changes or strain rate changes can be made without observable transients of the type predicted for the two extreme controlling mechanisms.

CONCLUSIONS

1. For conditions of continuously changing strain rate or Zener-Hollomon parameter the flow stresses of ferritic stainless steel and of Al-1% Mg depend only on the instantaneous values of strain and Zener-Hollomon parameter, i.e. these materials follow a mechanical equation of state when the rate of change of Zener-Hollomon parameter is below some critical value.

2. From the present experiments, in which the changes in strain rate were imposed mainly at constant values of $H = d \ln \dot{\epsilon} / d\epsilon$, the critical value of H for Al-1% Mg is $\gtrsim 45$ and for ferritic stainless steel is $\gtrsim 110$.

3. Commercial aluminium shows systematic deviations from a mechanical equation of state for all the experimental values of H . If a critical value of H exists for this material, it is very low $\cong 2$.

4. At higher values of H for commercial aluminium, the departure from an equation of state increases systematically with the value of H , irrespective of its sign, to approach an asymptotic situation at high values of H which is consistent with previous observations made with instantaneous changes in applied stress or strain rate.

5. The difference in behaviour between the experimental alloys and commercially pure metal is consistent with previously observed trends for other alloys and pure metals.

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REFERENCES

1. C. M. Sellars and W. J. McG. Tegart, *Int. Metall. Rev.* **17**, 1 (1972).
2. G. C. Cornfield and R. H. Johnson, *J. Iron Steel Inst.* **211**, 567 (1973).
3. R. A. Harding, Ph.D. thesis, Univ. of Sheffield (1976).
4. J-P. A. Immarrigeon and J. J. Jonas, *Acta metall.* **19**, 1053 (1971).
5. C. M. Young, S. L. Robinson and O. D. Sherby, *Acta metall.* **23**, 633 (1975).
6. D. R. Barraclough and C. M. Sellars, *Inst. of Physics, Conf. Ser.* No. 21, 111 (1976).
7. C. M. Sellars, J. P. Sah, J. Beynon and S. R. Foster, Plane-strain compression testing at elevated temperatures. Report to SRC, Univ. of Sheffield (1976).
8. J. P. Sah and C. M. Sellars, in *Hot Working and Forming Processes* (edited by C. M. Sellars and G. J. Davies), p. 62. Metals Soc., Lond. (1980).
9. S. R. Foster, Ph.D. thesis, Univ. of Sheffield (1981).
10. R. Lombry, C. Rossard and B. J. Thomas, *Rev. Metall. CIT* **79**, 975 (1981).
11. C. M. Young and O. D. Sherby, *J. Iron Steel Inst.* **211**, 640 (1973).
12. C. M. Sellars and W. J. McG. Tegart, *Mém. scient. Rev. Métall.* **63**, 731 (1966).
13. W. A. Wong and J. J. Jonas, *Trans. Am. Inst. Min. Engrs* **242**, 2271 (1968).
14. I. Gutierrez, Ph.D. thesis, Univ. of Navarra (1985).
15. E. S. Puchi, Ph.D. thesis, Univ. of Sheffield (1983).
16. R. Horiuchi and M. Otsuka, *Trans. Japan Inst. Metals* **13**, 284 (1974).
17. S. K. Mitra and D. McLean, *Proc. R. Soc.* **295**, 288 (1966).
18. I. D. Parker and B. Wilshire, *Phil. Mag.* **34**, 577 (1975).
19. B. L. Jones and C. M. Sellars, *Metal Sci. J.* **4**, 96 (1970).
20. G. Köning and W. Blum, *Acta metall.* **25**, 1531 (1977).
21. W. Blum and A. Finkel, *Acta metall.* **30**, 1705 (1982).
22. J. J. Urcola and C. M. Sellars, *Acta metall.* **35**, 2649 (1987).
23. J. J. Urcola and C. M. Sellars, *Acta metall.* **35**, 2659 (1987).