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HISTORY EFFECTS IN POLYCRYSTALLINE FCC METALS SUBJECTED TO RAPID CHANGES IN STRAIN RATE AND TEMPERATURE

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A review is presented of available experimental data on strain rate and temperature history effects for FCC polycrystalline metals together with new experimental results on copper and lead. The latter were obtained in experiments using a torsional split Hopkinson bar. The procedure involved incremental loading of the specimen, i.e. loading first at a low constant strain rate up to a predetermined initial shear strain at which a higher strain rate is superimposed without unloading the specimen. In all cases very prononunced strain rate history effects are observed; these are stronger in lead at room temperature than in aluminum or copper. It is evident from all the data presented that history effects play an important role which cannot be neglected in deriving constitutive relations to describe the plastic behavior of metals. It is shown in this paper that the influence of strain rate or temperature on the flow stress can be divided into two parts. The initial part is due to the existing work-hardened structure at that strain level, while the second is associated with the formation history of that structure. A possible explanation for these effects lies in dynamic recovery processes which take place during the slower deformation before the imposition of the rapid change in strain rate or temperature.

1. Introduction

The stress-strain behaviour of polycrystalline metals and alloys as well as single crystals is influenced both by the temperature at which deformation occurs and by the strain rate. This has been recognized for some time and, with this in mind, many experiments were performed to study material behaviour. In most of these experiments, however, or at least in those performed at high strain rates, the temperature is held fixed and a single strain rate is imposed during each test. While the results are of great value, they do not fully reflect the complications inherent in material behaviour. In particular, the influence of temperature history and of strain rate history are obscured, so that constitutive relations derived on the basis of such experiments are in a sense oversimplified.

In general, the current value of flow stress depends, among other things, on the history of strain rate and temperature; however, the nature of this dependence is not well understood at the present time. That there is an influence was clearly

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demonstrated by DORN, GOLDBERG and TIETZ (1948) in tests involving temperature changes during loading of aluminium at a constant strain rate. The same effect was observed by SYLWESTROWICZ (1958) in aluminum and in copper. Less pronounced effects were obtained by TIETZ and DORN (1949) for tests on aluminum at a constant temperature but with a change in strain rate from $3.6 \times 10^{-6}\ s^{-1}$ to $9.7 \times 10^{-4}\ s^{-1}$. KLEPACZKO (1968) studied strain rate history effects for polycrystalline aluminum (99.8% Al) and presented strain hardening curves obtained in shear tests at room temperature with a change in strain rate from $1.66 \times 10^{-5}\ s^{-1}$ to $0.624\ s^{-1}$. Strain rate history effects for aluminum were also observed by YOSHIDA and NAGATA (1966). FRANTZ and DUFFY (1972a) presented curves showing the dynamic stress-strain behaviour in shear of 1100-0 aluminum subjected to a sharp increase in strain rate from $10^{-5}\ s^{-1}$ to $10^{3}\ s^{-1}$ imposed at several strains in the range 0.01 to 0.15. Their experiments were performed with the aid of a modified split Hopkinson torsional bar. Additional data have also been reported for two kinds of aluminum by NICHOLAS (1971). Nearly all the above experiments were performed with aluminum so that very little data are available for other FCC metals. It is evident that there is a growing interest today in temperature history and strain-rate history effects. On the macroscopic side, and certainly as far as polycrystalline metals are concerned, these effects play an important role which cannot be neglected in deriving constitutive relations. There is evidence also that strain rate history effects are of consequence in the propagation of incremental elastic-plastic waves. This evidence is contained in the experimental results on aluminum and copper by CAMPBELL and DOWLING (1970) and by KLEPACZKO (1973) as well as for copper by YEW and RICHARDSON (1969), by CONVERY and PUGH (1970), and also by SANTOSHAM and RAMSEY (1970). A discussion of strain rate history effects from the point of view of incremental wave experiments has been published by KLEPACZKO (1972).

Finally, strain rate history effects may be involved in other aspects of material behaviour. For example, it has been shown by HAMSTAD and MUKHERJEE (1973), that acoustic emissions from a 7075-T6 aluminum alloy seem to be affected by strain rate history.

The above review indicates that a more general analysis of strain rate history and temperature history effects for polycrystalline FCC metals is needed. Such an analysis is the main purpose of this paper; it is undertaken on the basis of the results of complementary experiments reported on elsewhere which involve rapid increments in strain rate during loading of copper and lead specimens, and of the past work on aluminum.

2. EVIDENCE OF STRAIN RATE AND TEMPERATURE HISTORY EFFECTS IN FCC METALS

Available experimental data indicate that strain hardening in polycrystalline aggregates is a highly complicated process; there is no simple theory describing all the observed effects. For polycrystalline FCC metals incremental tests typically provide results such as those shown schematically in Fig. 1. Shear components of stress, strain and strain rate are plotted, although similar results are obtained with normal components. These data represent results of macroscopic tests, but they must reflect the very complicated character of the dislocation processes in polycrys-

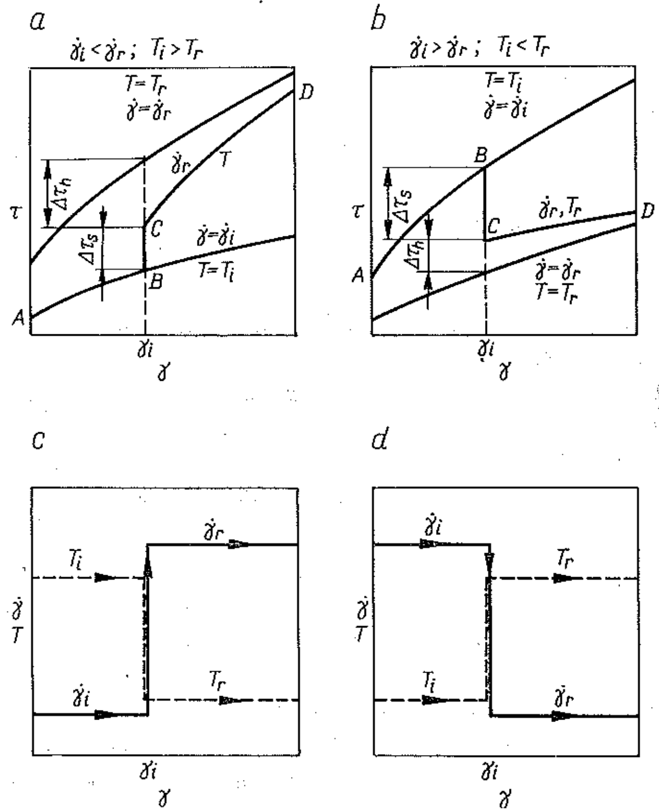

FIG. 1. Schematic diagrams showing behaviour typical of FCC metals; a) increase in strain rate from $\dot{\gamma}_i$ to $\dot{\gamma}_r$ or decrease in temperature from T_i to T_r; b) decrease in strain rate from $\dot{\gamma}_i$ to $\dot{\gamma}_r$ or increase in temperature from T_i to T_r.

tals. The lowest curve in Fig. 1 (a) represents a portion of the stress-strain relation for an FCC metal obtained during loading at a relatively low constant strain rate, e.g. $\dot{\gamma}_i = 10^{-4}\, s^{-1}$ or $10^{-5}\, s^{-1}$, while the upper curve is for a high constant strain rate, e.g. $\dot{\gamma}_r = 10^2\, s^{-1}$ or $10^3\, s^{-1}$. A specimen can also be loaded at one strain rate $\dot{\gamma}_i$ up to a strain γ_i at which point the strain rate is increased suddenly to $\dot{\gamma}_r$. Such a strain rate history is shown in Fig. 1 (c) plotted against strain. It results in the stress--strain curve in Fig. 1 (a) which follows the path *ABCD*. A similar path is followed when a specimen is loaded at a single constant strain rate if it suffers a sharp drop in temperature at γ_i. Typically, *BC* represents about half the distance between the two curves which are obtained by keeping both strain rate and temperature constant; furthermore, *CD* is a curve which gradually attains the upper stress-strain curve. The deformation along BC apparently occurs elastically, judging from the slope of the curve obtained during experiments, but the corresponding increment in stress,

$\Delta\tau_s$, is not sufficient to attain the upper curve. If $\Delta\tau$ denotes the difference at a given value of strain between the two flow stresses obtained at originally constant strain rates $\dot{\gamma}_i$ and $\dot{\gamma}_r$ (or constant temperatures T_i and T_r), then,

$$\Delta\tau = \Delta\tau_s + \Delta\tau_h. \tag{2.1}$$

Thus, $\Delta\tau_h$ is the difference at the same strain between the two stress-strain curves obtained at the same strain rate or temperature (see Fig. 1a). This difference must be due to the effects of the history of deformation since, following the increment at γ_i, the strain rate (or temperature) is the same along the two curves . In addition, the tangent modulus or strain hardening rate for the incremental curve, defined as $(d\tau/d\gamma)_r$, is higher than might be anticipated from the strain hardening rate at the same original strain rate of $\dot{\gamma}_r$.

For the case shown in Fig. 1 (b) the inverse phenomenon is observed. The decrease of strain rate from $\dot{\gamma}_i$ to $\dot{\gamma}_r$ or the temperature increase from T_i to T_r at a certain strain level γ_i results in a drop in the flow stress, again labelled $\Delta\tau_s$. The strain hardening rate $(d\tau/d\gamma)_r$ after the change of strain rate or temperature is lower than expected from the originally constant low strain rate $\dot{\gamma}_i$ or the higher temperature T_r. The relation (2.1) is again valid, although the quantities are now negative.

When strain rate (or temperature) is changed during loading, the observed history effects are probably due to the differences in the physical mechanisms of strain hardening operating at different strain rates (or temperatures). The incremental tests which are discussed in the present paper demonstrate the existence of these differences. More precisely, the influence of strain rate or temperature on the flow stress may be divided into two parts The initial part is due to the existing work-hardened structure at the strain level, while the second is associated with the formation history of that structure. Thus, the total stress difference $\Delta\tau$ between two stress-strain curves, each obtained at a constant strain rate (or temperature), can be divided into two parts. The first, denoted by $\Delta\tau_s$, is probably developed by a single thermally-activated dislocation mechanism, which is then dominant; experimentally, a measure of $\Delta\tau_s$ can be obtained by a sudden increase or decrease in strain rate (or temperature) since then there is essentially no change in structure. The second part, denoted by $\Delta\tau_h$, is related to the two different structures at the same value of strain. The difference between these two structures is probably due to a dynamic recovery process which occurred in the preceding deformation at the lower strain rate (or higher temperature).

It follows from the above that an incremental strain rate test is necessary in order to obtain a measure of what may be termed the real influence of strain rate on flow stress at a particular strain. The total difference commonly seen between two stress-strain curves obtained from tests performed at two different but constant strain rates is to a considerable extent due to existing microstructural differences. As a measure of the influence of strain rate, this total difference is therefore more apparent than real, and one can refer to $\Delta\tau_h$ as the apparent rate sensitivity and to $\Delta\tau_s$ as the real rate sensitivity. Thus, introducing the definition of rate sensitivity

$$\beta = \frac{\partial \tau}{\partial \ln \dot{\gamma}} \approx \frac{\Delta \tau}{\Delta \ln \dot{\gamma}} \tag{2.2}$$

and taking into account Eq. (2.1), one obtains

$$\beta = \frac{\Delta \tau_s}{\Delta \ln \dot{\gamma}} + \frac{\Delta \tau_h}{\Delta \ln \dot{\gamma}} \tag{2.3}$$

or

$$\beta = \beta_s + \beta_h . \tag{2.4}$$

The apparent rate sensitivity β_h deduced from constant strain rate tests must be attributed mainly to the difference in structure rather than to instantaneous strain rate response *per se.*

The main purpose of this paper is to review some existing results on strain rate and temperature history effects for FCC polycrystals. A second purpose is to report on new experimental results obtained in incremental type tests with copper and lead. Finally, using these data as a basis, a more general analysis of history effects for FCC polycrystalline metals is performed.

3. Experimental strain rate and temperature history data for aluminum

Early investigation of temperature history effects were devoted to the hypothesis of Ludwik (1909) that flow stress is a unique function of strain, temperature and strain rate. It is generally agreed that such a unique function frequently called the mechanical equation of state does not exist. A more detailed discussion of this topic is given by Lubahn and Felgar (1961) and by Conrad (1961).

As mentioned above, the early tests by Dorn, Goldberg and Tietz (1948) indicated the existence of temperature history effects for 1100-0 aluminum (formerly 2S-0). These results, reproduced in Fig. 2, show quite clearly that the temperature, during prior straining, influences the stress-strain diagram subsequent to a change in temperature. The schematic diagram in Fig. 1 (a) was drawn consistent with these results. The behaviour following an increase in temperature is also described in this paper and is shown schematically in Fig. 1 (b).

Considerable inprovements have been achieved in the last decade in the experimental techniques used to test metals under dynamic conditions. These have produced more reliable results especially for incremental type experiments in the high strain rate region. The most effective type of experiment is that performed under pure shear conditions, in which thin tubular specimens are used together with a modified torsional split Hopkinson bar.

An investigation of the behaviour of polycrystalline alluminum in pure shear using variable strain rates was that of Klepaczko (1968). In the first series of experiments a thin tubular specimen was deformed at a constant initial low strain rate

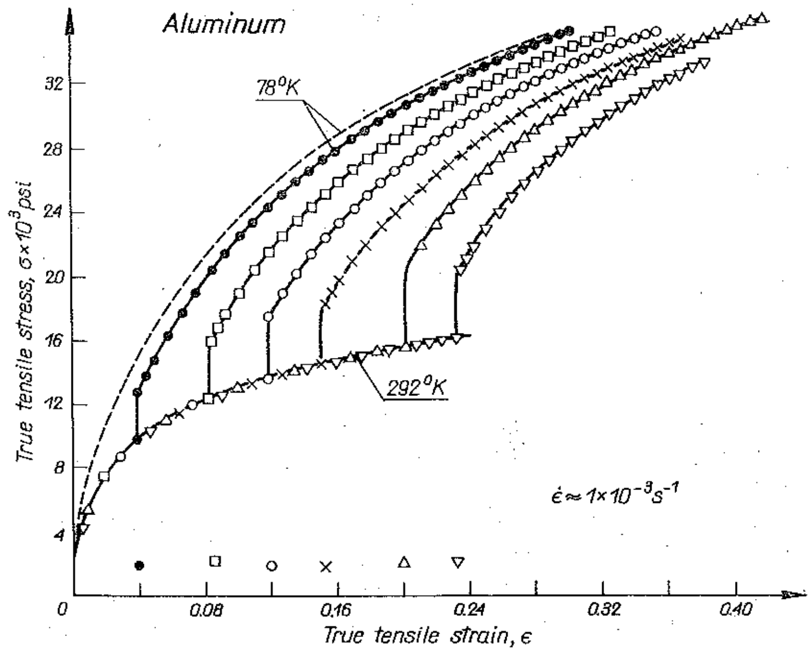

FIG. 2. Effect of temperature history on stress-strain behaviour of 1100-0 aluminum, after DORN, GOLDBERG and TIETZ (1948). Initial temperature $T_i=292°K$, reduced to $T_r=78°K$.

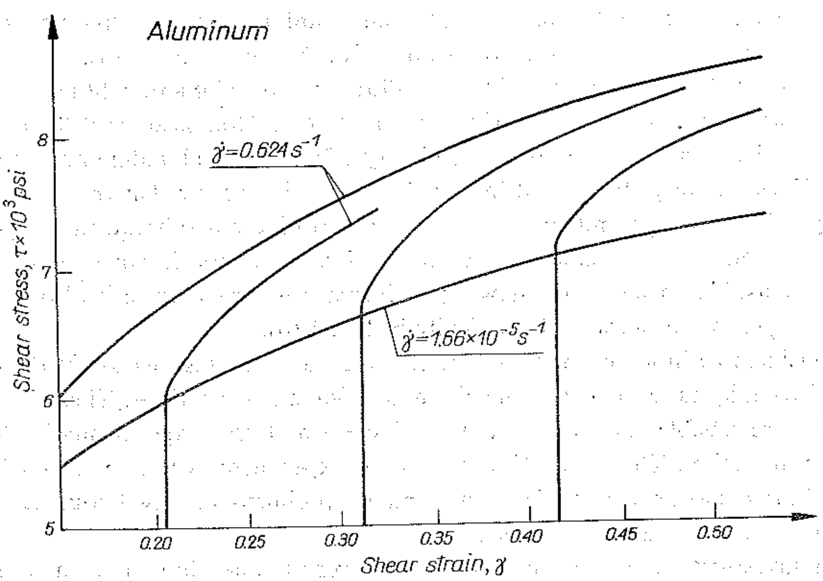

FIG. 3. Results of a strain rate change for three intitial values of strain γ_i; initial strain rate is $\dot{\gamma}_i = 1.66\times10^{-5}\,s^{-1}$, strain rate of reloading $\dot{\gamma}_r=0.624\,s^{-1}$, after KLEPACZKO (1968).

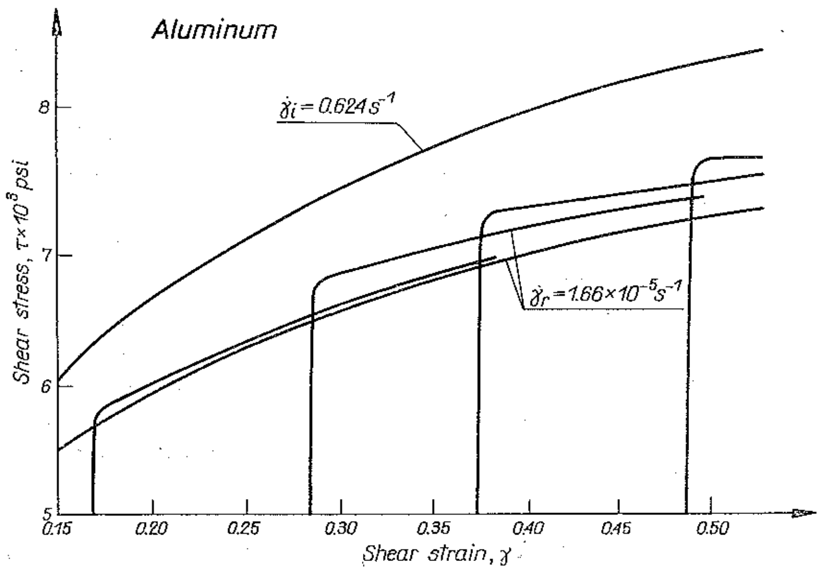

FIG. 4. Results of a strain rate change for four initial strains γ_i; initial strain rate is $\dot{\gamma}_i = 0.624\,s^{-1}$, and strain rate of reloading is $\dot{\gamma}_r = 1.66\times10^{-5}\,s^{-1}$, after KLEPACZKO (1968).

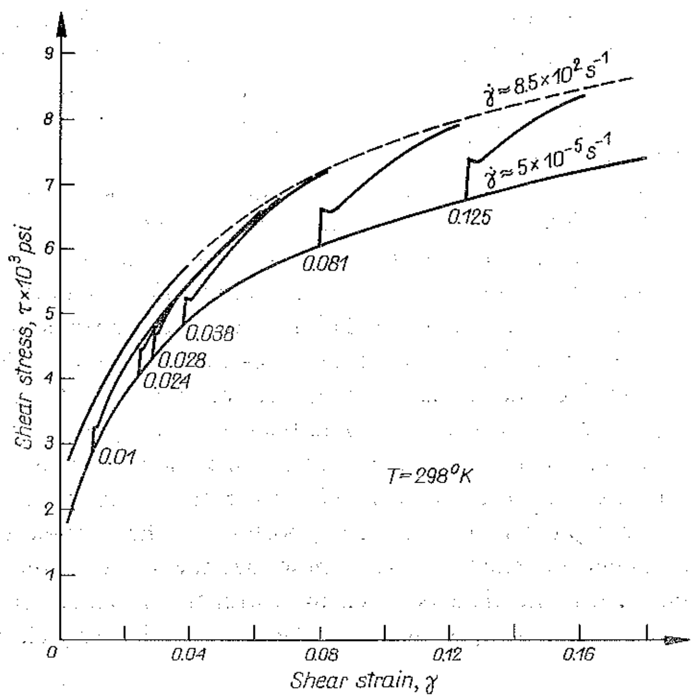

FIG. 5. Behaviour of 1100-0 aluminum in shear under a constant low strain rate, a constant high strain rate, and under incremental loading from $\dot{\gamma}_i = 5\times10^{-5}\,s^{-1}$ to $\dot{\gamma}_r = 8.5\times10^2\,s^{-1}$, after FRANTZ and DUFFY (1972a).

$\dot{\gamma}_i$ to an initial strain γ_i; it was then unloaded, and after an interval of about one minute reloaded at a higher constant strain rate $\dot{\gamma}_r$. Such experiments were repeated for several different values of the initial strain γ_i. The results are shown in Fig. 3. A second series of experiments was performed in the opposite manner, i.e. specimens were strained at a constant initial high strain rate $\dot{\gamma}_i$ up to different initial strains γ_i, and after an interval of about one minute reloaded at a lower constant strain rate $\dot{\gamma}_r$. The results are shown in Fig. 4.

From the results obtained thus far for aluminum, the following conclusions can be drawn. The behaviour observed with a temperature change is quite similar to that in Fig. 3 for a strain rate change in which $\dot{\gamma}_i < \dot{\gamma}_r$. Also, it is evident from a comparison of Fig. 3 and Fig. 4 that the values of the elastic stress response $\Delta\tau_s$ are smaller for $\dot{\gamma}_i < \dot{\gamma}_r$ than for $\dot{\gamma}_i > \dot{\gamma}_r$ at the same strain level. Assuming that the direction of the strain rate change does not influence $\Delta\tau_s$ increments, the differences of $\Delta\tau_s$ obtained at different $\dot{\gamma}_i$ must be attributed to different strain hardened states which are obtained with different strain rates at the same value of γ_i.

Until recently and except for tests in the creep range, all experimental investigations of strain rate history effects involved a complete unloading of the specimen and some lapse of time before a new strain rate was imposed by reloading. It is desirable to perform experiments in which unloading is eliminated and the new strain rate is superimposed directly on the initial one. The recent experiments of CAMPBELL and DOWLING (1970), NICHOLAS (1971) and FRANTZ and DUFFY (1972a) employed direct superposition of torsional loading. In the last mentioned paper all incremental experiments were performed by means of a modified torsional split Hopkinson bar in which a nominal strain rate $8.5 \times 10^2\ s^{-1}$ in shear is superimposed within a very short rise time with no unloading on a slow initial strain rate. The increment in strain rate is imposed at a previously selected value of shear strain γ_i within a range up to $\gamma_i = 0.15$. The results obtained by these authors for 1100-0 aluminum, shown in Fig. 5, are quite similar to those in Fig. 3 in spite of the higher incremental strain rate. A somparison of the strain and strain rate histories for the data in Fig. 3 and Fig. 5 is shown in Fig. 6. The only difference in the appearance of the stress-strain behaviour comes in the presence of an upper and lower yield stress-in the initial portion of the incremental stress-strain diagrams in Fig. 5. Otherwise, there is a considerable similarlity in the further portions of the incremental stressstrain diagrams. After a strain of about 0.005 (following the change in strain rate), the stress level of all incremental loading shown a smooth, steady rise gradually approaching the stress-strain curve at the higher constant strain rate. As expected, the strain hardening rate for the incremental portions of the stress-strain diagrams is very high. Results presented in Fig. 5 show that the shape of the incremental stress-strain curve near the yield point changes with the initial strain γ_i at which the increment is applied. As γ_i is increased, the stress level after yield drops more noticeably.

The question arises as to whether the drop in the stress level observed during the initial incremental response is due to the test procedure or to intrinsic material properties. In particular, there is a question, which has been discussed in more detail in the paper by NICHOLAS (1972), as to whether or not equilibrium is attained throughout the specimen during the initial part of the test. Some experimental evidence was obtained in favor of material behaviour; this is presented in FRANTZ

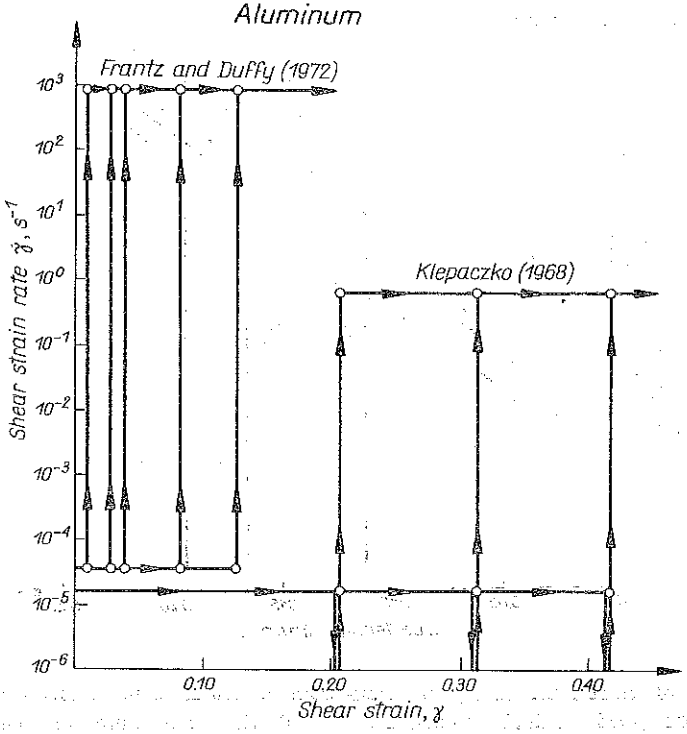

FIG. 6. Comparison of strain rate histories for stress-strain incremental curves of Fig. 3 and Fig. 5

and DUFFY (1972a). Also, it may be noted that the upper and lower yield points can be reproduced exactly for incremental conditions with the aid of the JOHNSTON--GILMAN approach (1959) if the concept of different generation rates of dislocations at different strain rates is assumed.

The schematic diagram in Fig. 1, which shows the incremental behaviour of a FCC metal, is in accordance with the incremental behaviour of aluminum reviewed above. By comparing the data in Fig. 5 with that in Fig. 3, it is seen that the influence of strain rate history is much the same in spite of the different strain and strain rate regions involved. It must be recognized then that strain rate and temperature history effects play an important role in the plastic behaviour of polycrystalline aluminum.

4. STRAIN RATE AND TEMPERATURE HISTORY DATA FOR COPPER

In contrast with aluminum there is almost a complete lack of data for polycrystalline copper. To the authors' knowledge the only contribution made so far concerning temperature history effects in copper is that due to SYLWESTROWICZ (1958). In his experiments tensile tests were performed at a constant strain rate of $\sim 10^{-3} s^{-1}$ at two different temperatures, 300°K and 76°K. Next, another series of specimens were strained at a temperature $T_i = 300$°K to four different values of initial strain

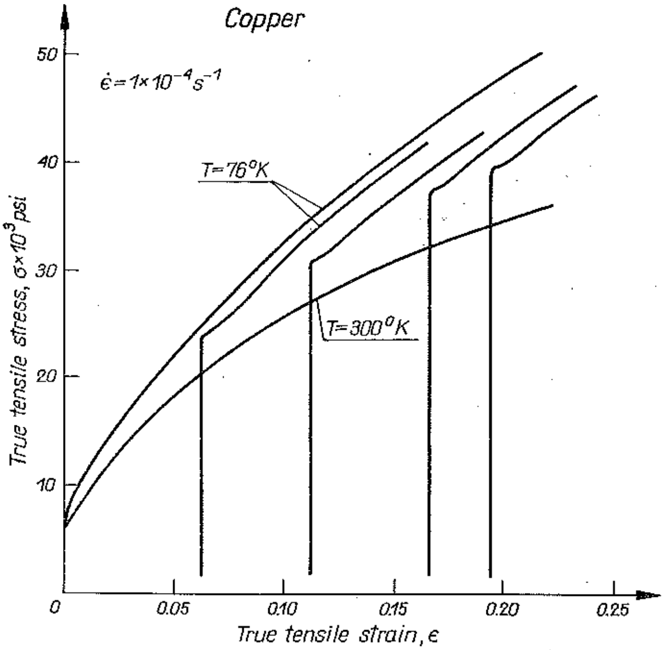

FIG. 7. Effect of temperature history on stress-strain behaviour of polycrystalline copper, after SYLWESTROWICZ (1958). Initial temperature $T_i = 300$°K, temperature of reloading 76°K; $\dot{\varepsilon} = 1 \times 10^{-4} s^{-1}$.

$\varepsilon_i = 0.06$, 0.11, 0.165, 0.195. A partial unloading occurred and the temperature was changed to $T_r = 76$°K; straining then continued at this last temperature. The second series of experiments involved increments imposed in the oposite direction; specimens were prestrained at $T_i = 76$°K up to six different values of initial strain where the temperature was increased to $T_r = 300$°K and partial unloading occurred then, straining continued at this higher temperature. These data, obtained by Sylwestrowicz for polycrystalline copper are shown schematically in Fig. 7 and agree with the expected behaviour shown in Fig. 1. The temperature history effects are clearly visible, but are much less dramatic than those shown in Fig. 2 for aluminum. It may be noted here that the temperature of 300°K constitutes a smaller fraction of the homologous temperature for copper (≈ 0.32).

Since the effect of increments in strain rate on the shape of stress-strain diagram for copper was unknown, new incremental strain rate tests were undertaken on annealed copper, KLEPACZKO (1974). Two series of tests were performed. In the first, incremental tensile tests were performed using a specially instrumented MTS universal testing machine. The specimen was first strained at a constant strain rate $\dot{\varepsilon}_i = 2.2 \times 10^{-5}\ s^{-1}$ up to an initial strain $\varepsilon_i = 0.15$, at which point the strain rate was increased to $\dot{\varepsilon}_r = 0.63\ s^{-1}$. The initial as well as the incremental portions of the stress-strain diagram were recorded. The incremental part of the test was recorded as a function of time with the aid of a fastgalvanometer oscillograph.

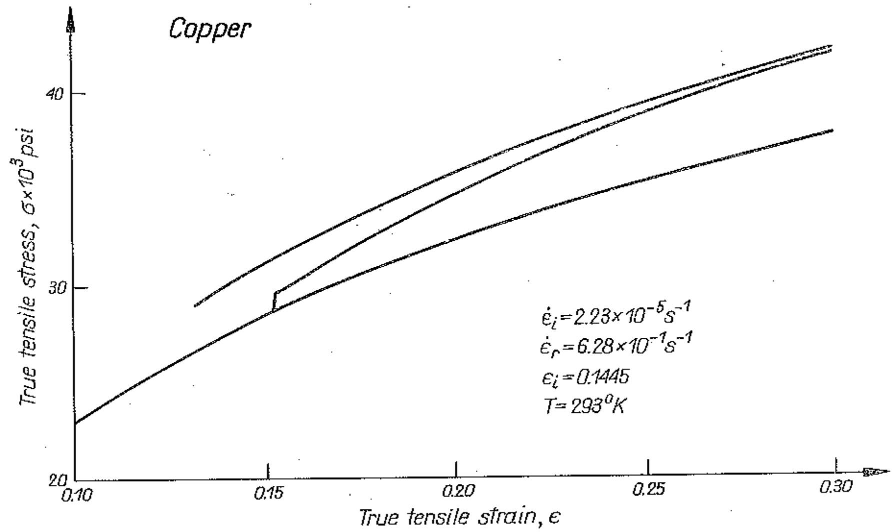

FIG. 8. Result of a strain rate change during a tensile test on polycrystalline copper; $\varepsilon_i \approx 0.153$, $\varepsilon_i = 2.22 \times 10^{-5}\ s^{-1}$, $\varepsilon_r = 0.53\ s^{-1}$.

This type of oscillograph was used because of the low frequency response of the built-in $x-y$ recorder of the machine. In processing the data, time was eliminated from the traces of the tensile force $P(t)$ and of the grip displacement $\Delta l(t)$. This procedure provides a P-Δl diagram which enables us to obtain an incremental stress-strain curve. In addition, a series of constant strain rate tests were performed at the same two strain rates, $\dot{\varepsilon}_i$ and $\dot{\varepsilon}_r$. A representative result is shown in Fig. 8. Again the strain rate history effect is clear.

It should also be noted that the incremental portion of the stress-strain diagram in Fig. 8 is very similar to that of Fig. 7 obtained with a change of temperature from 300°K to 76°K. In both diagrams there is a tendency in the first portion of the incremental curves for some inflections to appear, i.e. the strain hardening rate in this region is not decreasing monotonically.

The second series of tests, and the more important one, employed basically the same experimental technique used by FRANTZ and DUFFY (1972a) for aluminum. Again, the modified torsional split Hopkinson bar is used and a nearly constant high shear strain rate is superimposed within a very short rise time and with no unloading on a low constant strain rate. The incremental loading is imposed at the following five previously selected values of γ_i: 0.05; 0.10; 0.15; 0.20 and 0.25. For each value of γ_i at least three tests were completed. The measurements of reflected and transmitted pulses from the oscillograms were made with the aid of a computer program which gives the incremental portion of the stress-strain diagrams, as well as strain rate as a function of shear strain. The final results of these experiments on polycrystalline copper are shown in Fig. 9. Each incremental curve on this diagram represents the average of at least three tests. These experimental results are shown partly as dotted lines to indicate that the strain rate is not constant during

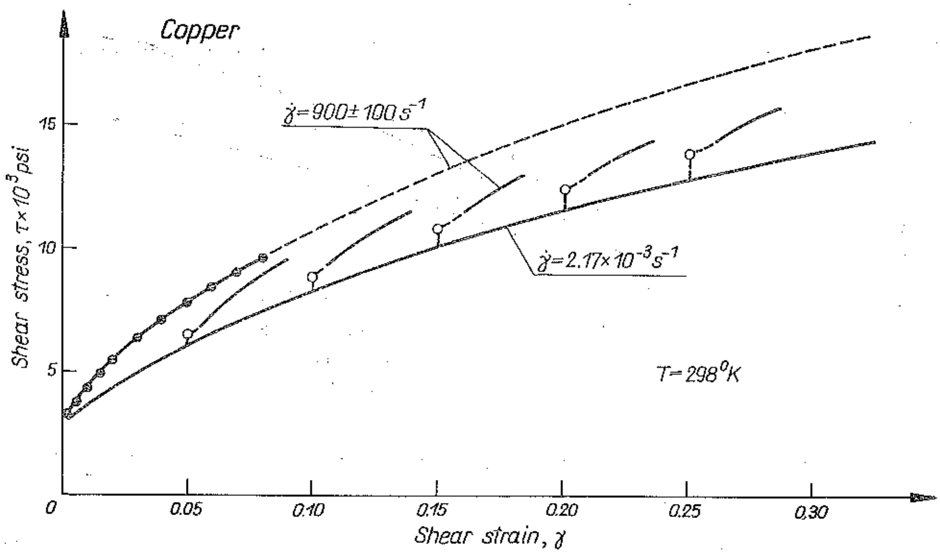

FIG. 9. Result of a strain rate change for five initial shear strains for copper, initial strain rate $\dot{\gamma}_i=2.17\times 10^{-3}\,s^{-1}$, strain rate of incremental loading $\dot{\gamma}_r\times 900\ s^{-1}$.

the entire incremental test: during the initial part of the incremental curve which is shown dotted it is increasing nearly linearly to the final maximum value. The solid line indicates that strain rate has attained a value of about $900\pm 100\ s^{-1}$ during that portion of the incremental test. Although the split Hopkinson bar technique is not adequate to measure exactly the initial response of the specimen, these responses appear to be elastic with the slope near that of the shear modulus G. The end of this elastic behaviour where plastic deformation first takes place was estimated as an averaged value from three tests. The average value of strain rate which is reached at these points is about $350\ s^{-1}$. After this point is reached plastic flow occurs with increasing strain while the strain rate increases from $350\ s^{-1}$ to about $900\ s^{-1}$, as indicated by the dotted line. This fact may be one of the factors causing the small inflection which is observed in the initial part of each incremental curve. The last portions of the incremental curves, denoted by solid lines, agree with the schematic pictures in Fig. 1 (a).

Comparison of the results in Fig. 7, which were obtained as the result of changes of temperature from 300°K to 76°K, with those in Fig. 9, which resulted from changes of strain rate from $2.17\times 10^{-3}\ s^{-1}$ to $900\ s^{-1}$, reveals a great similarity in all aspects of the stress-strain behaviour. This conclusion applies even to the inflections which appear in both figures in the initial portions of the incremental curves. Such an observation indicates that in the case of copper the dislocation processes responsible for temperature history and strain rate history effects are of the same nature.

5. Strain rate history data for lead

Lead at room temperature is a material with a reputation for being highly strain--rate sensitive. In view of this, a series of incremental tests was performed whose principal purpose was to obtain additional information concerning the effects of strain rate and strain rate history on the mechanical behaviour of lead. Essentially, the same experimental technique was applied as in the incremental testing of aluminum and copper. The only difference was in the shape of the specimens. Lead specimens were prepared in the shape of a ring, with a gage length and a wall thickness each fo 0.100 inch. Since lead is soft enough to deform plastically during handling, the specimens were first blanked out and then cemented to an aluminum mounted tube with the same cross-sectional dimensions as the Hopkinson tube. After the cement had cured, the specimens were cut to the proper length. The completed assembly was then cemented to the other end of the Hopkinson tube.

In these tests, a specimen being deformed at a low strain rate in shear is subjected suddenly and with no unloading to a dynamic rate: the difference between low and high strain rates it about seven orders of magnitude. The material response to the strain rate increase is recorded and analyzed in the usual way for the split Hopkinson bar technique. A more precise description of the experiments on lead is given in the report by Frantz and Duffy (1972b).

Blocks of lead of 99.99% purity were used to make the specimens. In order to reduce the grain size and provide a more homogeneous structure and more nearly isotropic properties, the blocks of lead were compressed in a static loading machine. This compression was done alternately in each of two orthogonal directions up to strain of 35% in each direction. Specimens were machined from the blocks following this preliminary compression process. Immediately after machining, the specimens were refrigerated at about 5°C and kept at this temperature until testing; several of them were examined microscopically in the course of the testing program, but no grain growth could be detected. There was about 20 grains in the wall thickness of these specimens.

The results of the entire series of incremental tests are summarized by the data presented in Fig. 10. Also included are the average stress-strain curves representing the behaviour observed during loading entirely at each of the strain rates, i.e. $\dot{\gamma}=0.7\times\times 10^{-4}\ s^{-1}$ and $\dot{\gamma}=10^{3}\ s^{-1}$. The stress-strain curves corresponding to the two low strain rates represent the averages of all data obtained from the static portions of the appropriate incremental test. The stress-strain curve representing loading at an entirely high strain rate is the average of two tests. These data agree very well with the results of earlier tests by Duffy, Hawley, and Frantz (1972). In presenting the results of Fig. 10, the incremental stress-strain curves are plotted only up to the point at which the strain rate first starts to decrease, i.e. only for the constant rate portion of the test.

The initial incremental response shows a stress increase at a slope approximately that of the shear modulus G; it should be noted that this stress increase occurs simultaneously with the strain rate increase. At first sight, this initial elastic response

$\Delta\tau_s$ may seen rather small for such a strongly rate dependent material. However, this elastic increase in stress for lead is about 15% to 20% of the existing low-strain--rate flow stress at that strain. Therefore, it is proportionally greater than for aluminum, Fig. 5, or copper, Fig. 9. On the other hand, a comparison of $\Delta\tau_s$ to $\Delta\tau_h$

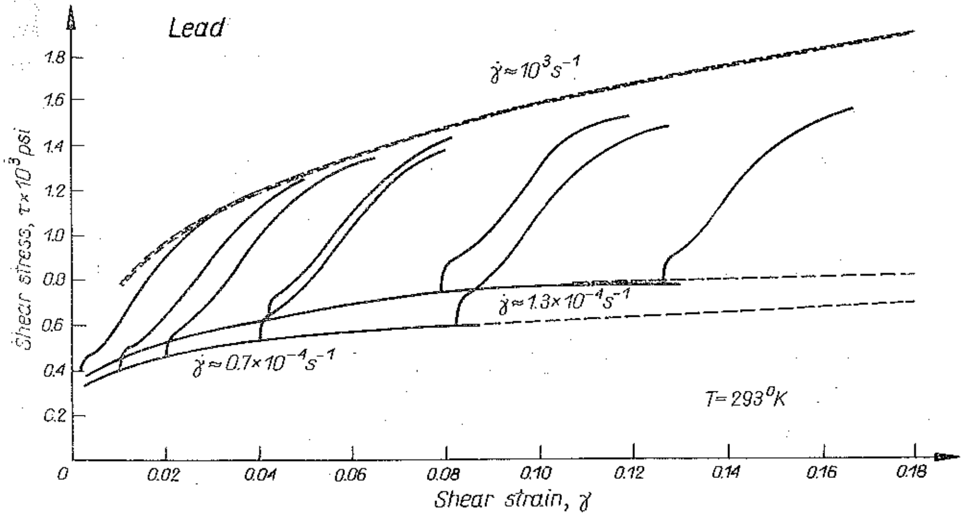

FIG. 10. Behaviour of 99.99% pure lead in shear under a constant low-strain-rate, a constant high strain rate, and under incremental loading from $\dot{\gamma}_i = 0.7 \times 10^{-4}\ s^{-1}$ and $\dot{\gamma}_i = 1.3 \times 10^{-4}\ s^{-1}$ to $\dot{\gamma}_r \approx 10^3\ s^{-1}$.

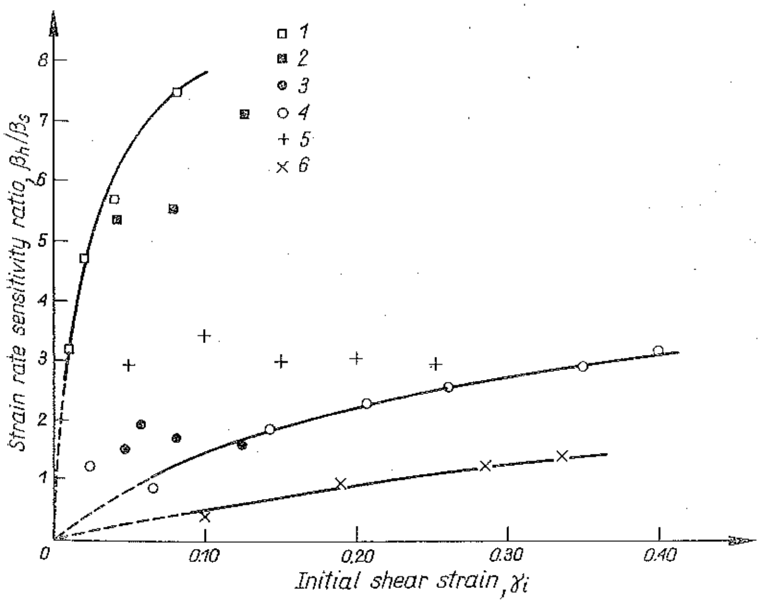

FIG. 11. Ratio of the rate sensitivities β_h/β_r as measured from the ratio of the stress increments $\Delta\tau_h/\Delta\tau_s$; data based on the strain rate and temperature incremental tests for aluminum, copper and lead.

at γ_i shows that the estimated ratio of $\Delta\tau_h/\Delta\tau_s$ is much higher for lead than for aluminum or copper. It should also be noted that for lead the nearly identical increments of $\Delta\tau_s$ are generated at nominal initial strains γ_i from 0.01 through 0.125.

A third important feature of the present results is the slope of the stress-strain curves during deformation at the rate of 10^3 s^{-1}. The stress-strain curve at a constant strain rate of 10^3 s^{-1} and all incremental curves are concave up at strains ranging from initial yield up to 0.020 to 0.025 after the addition of the incremental load. This behaviour had been observed in earlier tests on lead at a constant strain rate of 10^3 s^{-1} (but was not discernable at rates of 5×10^3 s^{-1} to 8×10^3 s^{-1}). NICHOLAS (1972) offered the explanation that the concave-upward stress-strain behaviour may result from a lack of equilibrium in the specimen during early stages of deformation. Several computations were made by FRANTZ and DUFFY (1972) to investigate this possibility. These calculations indicate that this concave up freature was unlikely to be due to a lack of equilibrium in the specimen. However, lack of equilibrium cannot be eliminated as the explanation and, for certain materials and loading conditions, probably applies within certain ranges of the dynamic response of the specimen. In the incremental tests a more probable explanation of the behaviour is offered by the fact that all elastic responses for lead take place with very low incremental stress values. This fact implies that yielding occurs within a time interval which is less than the rise time t_r of the incident pulse. Usually, 10 μs$<t_r<$15 μs. Immediately after yield the stress-strain curve levels off and this is probably due partly to intrinsic material behaviour as well as to a non-constancy of the strain rate at the early time of deformation. After that the curve rises sharply indicating a high rate of strain hardening. This steep rate of hardening continues up to t_r the instant when the strain rate levels off to a constant value. When the strain rate reaches its constant value, it can be expected that the strain hardening rate will take on a value characteristic of lead at that stage of straining. Further experimental work is needed to establish the role of intrinsic material behaviour in the observed stress-strain diagram of lead. In this regard, experiments involving changes in strain rate at different temperatures will be very important.

6. DISCUSSION

The most important general observation which can be made on the basis of the experimental results presented above is that strain rate and temperature history effects play an important role in the plastic behaviour of polycrystalline FCC metals. Possible explanations for these effects lie in the dynamic recovery dislocation processes and in the processes usually responsible for creep at intermediate temperatures and high stresses. A discussion of factors which may be responsible for temperature and rate-history effects is presented below without attempting to identify the particular dislocation processes involved. The latter will form a succeeding step in the analysis of strain rate history and temperature history effects in FCC metals.

During plastic straining of a metal its internal structure changes: strain hardening takes place due to an increasing dislocation density and, simultaneously, some softening occurs due to dynamic recovery. These are competing tendencies which together produce an effective strain hardening rate. If the specimen is strained at a constant strain rate and temperature, then, these two rates (hardening and softening) are characteristic of the material in the current state and this situation does not change until either strain rate or temperature is changed. Thus the effective hardening rate is the result of microscopic mechanisms some of which generate dislocations and others annihilate them. This is frequently expressed in terms of the tangent modulus of the stress-strain diagram. For example, for nonisothermal conditions

$$\frac{d\tau}{d\gamma}=\left.\frac{\partial\tau}{\partial\gamma}\right|_{t,T}+\frac{1}{\dot{\gamma}}\left.\frac{\partial\tau}{\partial t}\right|_{\gamma,T}+\xi\left.\frac{\partial\tau}{\partial T}\right|_{\gamma,T}, \tag{6.1}$$

where t is time of deformation, T is absolute temperature, and $\xi=dT/d\gamma$. The second term in Eq. (6.1) is responsible for the dynamic recovery contribution to the rate of hardening and $\partial\tau/\partial t|_{\gamma,T}$ is negative. It is also evident that dynamic recovery is part of the total strain rate sensitivity effect and is more important the more time elapses. Hence, at lower strain rates the seocnd term in Eq. (6.1) is more important. When the strain rate is high, on the other hand, the contribution of the second term is negligible. The third term is positive for a decrease in temperature during straining, since the partial derivative $\partial\tau/\partial T|_{\gamma,t}$ is usually negative. Relation (6.1) allows for a quantitative description of strain rate and temperature history effects when the magnitudes of the two partial derivatives, $\partial\tau/\partial t|_{\gamma,T}$ and $\partial\tau/\partial T|_{\gamma,t}$ are known. For purposes of this paper it provides a qualitative description of the effects involved.

In studying the reasons for strain rate sensitivity the most fundamental question which arises is the extent to which the effective strain rate sensitivity is due to dynamic recovery (i.e. dislocation annihilation mechanisms) or to low temperature thermally-activated dislocation processes. The answer to this question cannot be obtained from tests performed at constant strain rates or temperatures; it requires an analysis of results of incremental tests.

Following the definition of strain rate sensitivity (2.2) a similar definition of temperature sensitivity can also be introduced

$$\beta_T=\frac{\partial\tau}{\partial T}=\frac{\Delta\tau}{\Delta T}. \tag{6.2}$$

As was done above for the strain rate sensitivity this can be divided into components β_s and β_h. Thus, Eq. (2.4) still holds, and can be rewritten

$$\beta=\beta_s\left(1+\frac{\beta_h}{\beta_s}\right), \tag{6.3}$$

where, in our case $\beta_h/\beta_s=\Delta\tau_h/\Delta\tau_s$. This last ratio holds for either an increment $\Delta\ln\dot{\gamma}$ or ΔT as long as this increment is of the same magnitude as the difference in the constant strain rates or temperatures. As a result Eq. (6.3) can be rewritten in the form

$$\frac{\beta}{\beta_s}=1+\frac{\Delta\tau_h}{\Delta\tau_s}. \tag{6.4}$$

The ratio $\beta_h/\beta_s = \Delta\tau_h/\Delta\tau_s$ is a measure of the dynamic recovery contribution to the strain rate or temperature sensitivity developed by the currently operating thermally-activated dislocation mechanism. If β_h/β_s is equal to zero then there is no strain rate or temperature history effect and $\beta=\beta_s$. When the value of β_h/β_s is large, then, of course, the contribution of the low-temperature thermally-activated dislocation mechanisms are less significant. To estimate the importance of this contribution, values of β_h/β_s were computed for different initial strains, γ_i. The experimental results used for this computation were taken from Figs. 2 and 5 for aluminum, Figs. 7 and 9 for copper, and Fig. 10 for lead. Some of these data are based on results of tensile tests, and they were converted to shear by the application of the Huber--Mises yield condition. It may be mentioned here that for all these tests the initial temperature is $T_i=293°\mathrm{K}$ and that the initial strain rate, $\dot{\gamma}_i$, ranges from $10^{-5}\ s^{-1}$ to $10^{-3}\ s^{-1}$. The computed values of β_h/β_s are shown in Fig. 11. The highest values are obtained with lead; for instance, at a shear strain $\gamma_i\approx 0.10$, β_h/β_s is approximately equal to 8. This means, as expected, that at room temperature the dynamic recovery contribution is more important for this metal than for aluminum and copper. It also means that the principal contribution to strain rate sensitivity in lead at room temperature and low strain rates is due to dislocation mechanisms associated with creep (dynamic recovery). The three points shown as black squares are slightly lower in the graph; they were obtained at a higher initial strain rate. Thus, in the case of lead the ratio β_h/β_s increases sharply for increasing initial strains, which indicates that during initial deformation at a low strain rate the rate of dynamic recovery is very intense.

The lower solid line are obtained from the incremental temperature tests for aluminum and copper. Again, as expected, the rate of dynamic recovery at room temperature and at a low strain rate is less intense for copper than for aluminum, but the values of β_h/β_s are still far above zero. The results obtained in strain rate incremental tests are less conclusive since the values of β_h/β_s are nearly constant; for aluminum about 1.5 and for copper about 3. These values are relatively high compared to the remainder of the data.

When considered as a whole the above results indicate that most of the strain rate sensitivity at room temperature and at constant low strain rates is due to the stress relaxation process developed by dynamic recovery, which leads to a dislocation annihilation, rather than being due to low temperature thermally-activated dislocation mechanisms.

The present tests with the split Hopkinson bar provided stress-strain curves for strains subsequent to the application of the strain rate increment up to about 0.04 for aluminum and lead and about 0.08 for copper. In order to study history effects at larger strains these curves were extended according to the following definition of strain rate sensitivity,

$$n=\left(\frac{\partial \log \tau}{\partial \log \dot{\gamma}}\right)_{T,\gamma}. \tag{6.5}$$

This definition is more useful for this purpose than the definition of β used in Eq (2.2) since it is less sensitive to strain levels. The calculations to extend the curve were performed as follows: first, the values of n were calculated for different strains within the region of strain obtained from the split Hopkinson apparatus using low strain rate and high strain rate portions of the stress-strain diagrams. These averaged values of n are: $n_{Al}=0.0105$; $n_{Cu}=0.0201$; and $n_{Pb}=0.0417$. The calculations showed that the values of n are almost strain independent. Next, the further portions of the dynamic stress-strain relations were estimated from the formula

$$\tau_1=\tau_2\left(\frac{\dot{\gamma}_r}{\dot{\gamma}_i}\right)^n. \tag{6.6}$$

Estimated extension of the stress-strain curves for constant strain rates are indicated by the dotted lines in Fig. 5, Fig. 9 and Fig. 10. On the basis of these extensions the values of β_h were estimated for strains greater than those reached with the split Hopkinson bar technique.

7. Conclusions

(i) Incremental tests, whether the increments be in strain rate or temperature, show that strain rate history and temperature history effects play a very important role in the plastic deformation of FCC polycrystalline metals at room temperature and at low initial strain rates.

(ii) The total stress difference $\Delta\tau$ between two stress-strain curves each obtained at a different but constant strain rate (or temperature) can be divided at any strain level into two parts. The first part, $\Delta\tau_s$, is due to the thermally-activated response at one strain level of the existing work-hardened structure. It may be considered as the real strain rate sensitivity. The second part, $\Delta\tau_h$, is associated with the formation of a certain dislocation structure; thus this part is history dependent.

(iii) The strain rate sensitivity observed in tests at different but constant strain rates is due partially to the dynamic recovery which occurred during the preceding straining and which probably involves the annihilation of dislocations. This part of the strain rate sensitivity is given by $\beta_h=\Delta\tau_h/\Delta \ln\dot{\gamma}$, which can be evaluated by means of an incremental test (a test involving a sharp increment in strain rate). The remaining part of the strain rate sensitivity can be attributed to a thermally-activated dislocation mechanism. It can be evaluated on the basis of the incremental test by making use of $\beta_s=\Delta\tau_s/\Delta \ln\dot{\gamma}$.

(iv) On the basis of the experiments performed on lead one may conclude that the strain rate effect observed at two constant strain rates for this metal receives a much greater contribution from the dynamic recovery processes than from the thermally-activated processes operating in that range of strain rates.

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Streszczenie

EFEKTY HISTORII W POLIKRYSTALICZNYCH PŁASKOCENTRYCZNYCH METALACH PODDANYCH ZMIANOM PRĘDKOŚCI ODKSZTAŁCENIA I TEMPERATURY PODCZAS SZYBKIEGO OBCIĄŻENIA

Podano przegląd dostępnych w literaturze danych doświadczalnych na temat wpływu historii, prędkości odkształcenia i temperatury na zachowanie się polikrystalicznych metali o płaskocentrycznej strukturze sieci krystalicznej, z najnowszymi rezultatami dla miedzi i ołowiu włącznie, otrzymanymi na skrętnym pręcie Hopkinsona. Procedura doświadczeń polegała na przyrostowym obciążaniu próbki; tzn. najpierw obciążano próbkę przy niskiej stałej prędkości odkształcenia do z góry ustalonego początkowego stanu odkształcenia, na który następnie nakładano wyższą prędkość odkształcenia bez obciążania próbki. We wszystkich przypadkach zaobserwowano bardzo wyraźne efekty historii prędkości odkształcenia. Są one większe dla ołowiu w temperaturze pokojowej, niż dla aluminium lub miedzi.

Na podstawie wszystkich przytoczonych danych jest oczywiste, że efekty historii odgrywają istotną rolę i nie można ich zaniedbywać przy wyprowadzaniu związków konstytutywnych służących do opisu plastycznego zachowania się metali. Wykazano w tej pracy, że wpływ prędkości odkształcenia lub temperatury na naprężenie plastycznego płynięcia można podzielić na dwie części. Część początkową powstałą wskutek zaistniałej struktury wzmocnienia na danym poziomie odkształcenia, podczas gdy druga część łączy się z historią tworzenia się tej struktury. Wyjaśnienie tych efektów jest możliwe za pomocą dynamicznego procesu zdrowienia, jakie zachodzi podczas wolniejszej deformacji przed nałożeniem szybkiej zmiany prędkości odkształcenia lub temperatury.

Резюме

ЭФФЕКТЫ ИСТОРИИ В ПОЛИКРИСТАЛЛИЧЕСКИХ ПЛОСКОЦЕНТРИРОВАННЫХ МЕТАЛЛАХ ПОДВЕРГНУТЫХ ИЗМЕНЕНИЯМ СКОРОСТИ ДЕФОРМАЦИИ И ТЕМПЕРАТУРЫ ВО ВРЕМЯ БЫСТРОГО НАГРУЖЕНИЯ

Дается обозрение доступных в литературе экспериментальных данных на тему влияния истории, скорости деформации и температуры на поведение поликристаллических металлов, с плоскоцентрированной структурой кристаллической решетки, с новейшими результатами

для меди и свинца включительно, полученными на крутильном стержне Гопкинсона. Процедура экспериментов заключалась в приросте нагружения образца т.зн, сначала образец нагружался при низкой постоянной скорости деформации к заранее устанивленному начальному деформационному состоянию, на которое затем накладывается высшая скорость деформации без разгрузки образца. Во всех случаях наблюдались очень отчетливые эффекты истории скорости деформации. Они больше для свинца в комнатной температуре, чеж для алюминия или меди. На основе всех приведенных данных становится очевидным, что эффекты истории играют сучественную роль и не можно ними пренебрегать при выводе определяющих соотношений, служащих для описания пластического поведения металлов, В этой работе показано, что влияние скорости деформации или температуры на напряжение пластического течения можно расделить на две части. Начальная часть возникла вследствие существования структуры упрочнения на данном уровне деформации, в то время как вторая часть связана и историей образования этой структуры. Выяснение этих эффектов возможно при помощи динамического процесса возврата, который имеет место во время более медленной деформации перед наложением быстрого изменения скорости деформации или температуры.

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