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HOT-WORKING MECHANISMS

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PHYSICAL METALLURGY DIVISION

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HOT-WORKING MECHANISMS*

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ABSTRACT

Before introducing the results of hot-working experiments, the techniques and their limitations are described. Microstructural observations and mechanical measurements are discussed separately. The microstructure of hot-worked single-phase alloys consists of polygonized cells; the degree of polygonization increases with rise in temperature, decrease in strain rate and greater facility of dislocation cross-slip and climb. The mechanism thus appears to be dynamic recovery without recrystallization. Recrystallization in the course of deformation may occur in multi-phase alloys. It is shown that the most suitable empirical expression relating strain rate, temperature and flow stress can be interpreted in terms of a stress-aided, thermally-activated process.

INTRODUCTION

Whereas deformation at high temperatures is utilized extensively in industry because of the high malleability and the low flow stress, the mechanisms of hot working have been subjected to a very limited investigation compared to cold working, creep or alloy strengthening. This situation arose partly because the experimental techniques, which will be discussed, are quite demanding. In the past decade, however, the advance in understanding has been rapid thanks to the fund of knowledge of dislocation mechanisms established by studies of strain hardening, recovery, recrystallization and creep. The results of investigations up until the beginning of 1967 have been reviewed in an earlier paper by the author⁽¹⁾ and in the ASM Seminar on Ductility by Tegart⁽²⁾. The present paper will outline the highlights of the theory and indicate the developments and questions arising from new experimental results.

EXPERIMENTAL TECHNIQUES

Deformation Techniques

Probably the most important experimental conditions to achieve is a uniform logarithmic strain, ϵ , of the order of 2 to 10 ($\epsilon = \ln(l/l_0)$) at a constant strain rate, since hot deformation is very sensitive to that variable. It is important to attain a strain in the above range since within it a steady state pre-

vail in which at constant temperature and strain rate, the stress is constant.

The tensile test is not of great use since uniform strain ceases with the commencement of necking at a strain of about 0.3 and rupture occurs at a strain less than 0.8 which is below both that for steady-state and that commonly achieved in industrial operations. In compression, friction between the specimen and the platens gives rise to non-uniform deformation which is exhibited as barrelling and which limits the strain to about 2 even with the best lubrication. Furthermore both tension and compression with constant crosshead-motion machines do not give deformation at constant strain rate. Constant strain rate in compression is attained on the cam plasto-meter in which the compressing cam reduces the rate of compression as the height of the specimen decreases.

The torsion test with short gauge-length tubular specimens is the only satisfactory solution since it gives uniform strain at constant strain rate. With solid test specimens, the strain and strain rate vary from zero at the centre to a maximum at the surface. The torsion test offers the further advantages that complex stress states similar to those of shaping processes can be produced by simultaneous axial loading and that multipass operations can be simulated by programming halts and rate changes in the deformation⁽³⁾.

Industrial processes themselves suffer from non-uniformity of strain and strain rate: this makes them unsuitable for experiments and complicates the quantitative application of results from other tests.

Microstructural Examination

Microstructural examination of the deformed metal, which is essential for resolving the operating mechanisms, is not without difficulty. The primary problem is the prevention of structural modifications upon cessation of deformation, since, in some metals, they proceed very rapidly at the high test temperature. Quenching must therefore be very rapid and the deformation must be effected in a single pass. Optical microscopy does not satisfactorily reveal the substructure except in a few cases. X-ray diffraction is not always easily applied because of the mixed structure resulting from partial recrystallization after deformation has ceased. Electron microscopy,

which has been available for only a decade, has proved a very effective tool giving considerable new insight.

MICROSTRUCTURAL CHANGES AND MECHANISMS

Initial Hardening Stage

During the initial hardening, the dislocations accumulate into tangles which, by a strain of about 0.2, roughly outline cells. With further strain, the cells decrease in size and become more sharply defined^(4, 5). As the steady state is approached, the rate of hardening decreases as the cells reach a terminal size which depends on temperature and strain rate. The limiting cell size increases gradually as the temperature of deformation is raised and is associated with a decreasing maximum flow stress. As the temperature increases, the cell size becomes more dependent on the strain rate.

Under hot-working conditions the cells are usually polygonized, that is, the internal dislocation

density is low and the wall dislocations are arranged in neat networks. The substructure resembles that produced by recovery of cold-worked metals; however, at high temperatures and low strain rates, the degree of polygonization is usually much higher than can be attained by recovery after cold working because of the intervention of recrystallization. Recovery concurrent with deformation is known as dynamic recovery.

These structural modifications have been observed in almost all metals and alloys regardless of their crystal structure. The degree of polygonization is less for metals in which cross-slip and climb are inhibited by extension of the dislocations, such as face-centred-cubic alloys of low stacking fault energy.

Steady State Deformation: Dynamic Recovery Mechanism

In metals, in which dislocations may easily cross-glide and climb, the substructure developed during the hardening stage persists throughout the steady-state deformation (Fig. 1). Throughout steady-state, the subgrain (cell) size and misorientation remain constant⁽⁴⁻⁶⁾; thus the dislocation density remains constant and permits the deformation to proceed without increase in stress⁽⁷⁾. A similar substructural permanence is observed during steady-state creep⁽⁸⁾. The dislocation

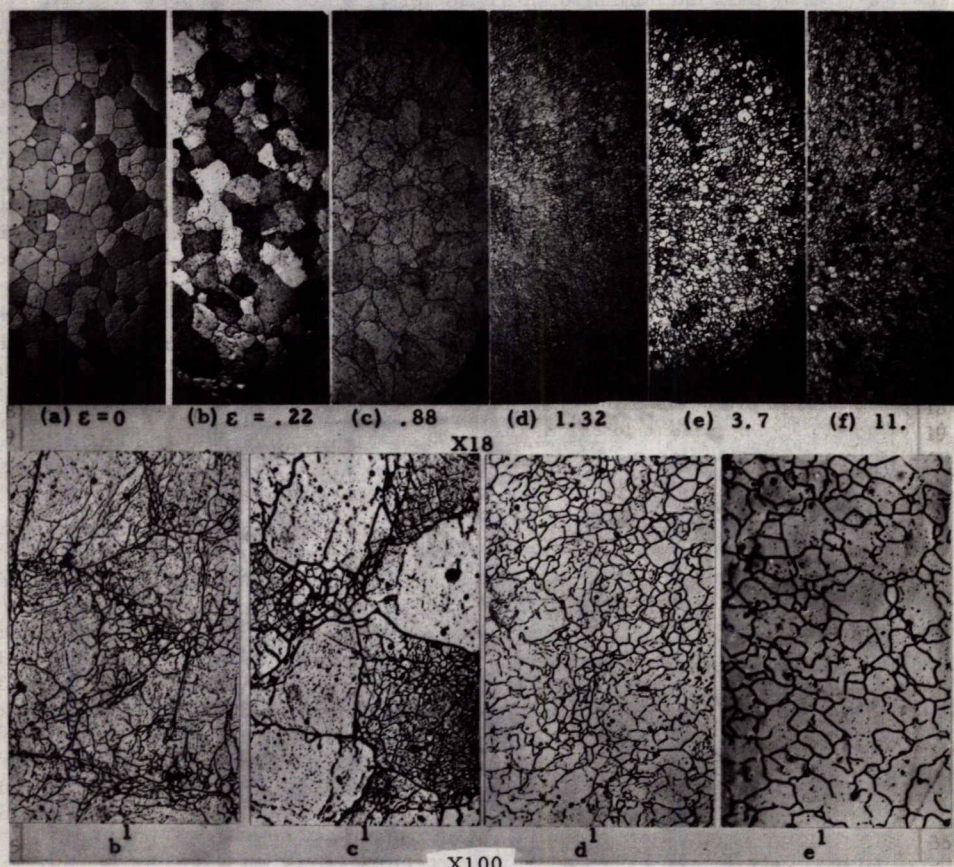


Figure 1 Microstructure of Fe-25% Cr after different true strains in torsion at 1100°C and at a strain rate of 1.1 sec⁻¹. Series a-f (X18) show approximately half of the torsion sample. Series b'-e' (X100) are from samples b-e respectively. Specimens were quenched in 1 sec. Rossard (6).

density remains constant because of dynamic recovery: the rate of annihilation, combination and egress at the grain boundaries equals the rate of dislocation generation. Since the subgrains remained equiaxed while undergoing extensive strain (however, the grains are elongated)^(4, 5, 7), it appears that the substructure is not a static, rigid framework but that the subgrains continuously break up and reform⁽⁷⁾. This process of repolygonization occurs as the boundary dislocations are annihilated or heavily jogged by the high flux of mobile dislocations⁽⁷⁾. Since the sub-boundaries observed are general dislocation boundaries of less than 5° misorientation, they are unable to migrate sufficiently rapidly to maintain equiaxiality. If the deformation is terminated at any strain in the steady state region, the continuously changing substructure always consists of equiaxed subgrains of constant average size and misorientation.

For different strain rates or temperatures, the balance between annihilation and generation is struck at a different dislocation density. As the number of thermally activated events per unit strain decreases with increase in strain rate, or decrease in temperature, the subgrain size decreases, the density of dislocations inside the cells increases and the walls become denser and more ragged (Fig. 2)⁽⁷⁾. The substructure in



Figure 2. Subgrain diameter and perfection, in commercial purity aluminum extruded at 445°C, decrease with strain rate: (a) 0.5 sec⁻¹ and (b) 5.5 sec⁻¹. Extrusion ratio 40:1.

secondary creep is affected in a similar manner by changes in temperature and creep rate⁽⁸⁾.

The investigation by Drube and Stüwe⁽⁵⁾ of the hot-worked structure of copper fully supports the above theory; however certain aspects are of special interest. Under optical examination, specimens deformed at high strain rates exhibit the original grains deformed, whereas those at low strain rates exhibit what appear to be new grains (the strain rate for the change in appearance increases as the temperature increases). After annealing for several minutes at the working temperature, high-strain-rate specimens take on a microstructure similar to that for low strain rates. Examination by transmission electron microscopy reveals a polygonized substructure with low misorientation between the cells regardless of optical microstructure; at the lower strain rate the subgrains are larger and sharper. No regions of abnormally low dislocation density are observed. The preferred orientation is observed to be the same for high-strain-rate, low-strain-rate, and annealed high-strain-rate specimens. The specimens exhibiting what appears to be new grains have not recrystallized; moreover, the observed microstructure differs from that of copper recrystallized after hot working, especially in the lack of annealing twins⁽⁹⁾.

In hot deformation, metals having different crystal structures but equal ease of dislocation cross-glide and climb have very similar substructures, e. g. aluminum⁽⁷⁾ and iron⁽¹⁰⁾. Metals having the same crystal structure but differing in the ability of the dislocations to move out of the slip plane, e. g. face-centred-cubic metals with different stacking fault energies, have considerably different structures (Fig. 3)⁽⁹⁾. To isolate the effect of stacking fault energy, the different metals must be deformed at a common level of thermal activation; deformation is therefore performed at the same homologous temperature (fraction of absolute melting temperature). Although the extent of recovery may be greatly reduced at low stacking fault energies (Cu 30% Zn, Cu 7% Ge) no recrystallization occurred during deformation^(5, 9). Dynamic recovery is usually diminished by the presence of impurities, as may be seen from comparison of Fig. 2 and Fig. 3a.

HOT DEFORMATION OF IRON ALLOYS

The microstructural evolution during the hot working of single-phase ferritic alloys (Fig. 1) is much the same as that in aluminum^(6, 10-15). The resulting subgrains are similar to those formed during creep^(8, 16) or recovery⁽¹⁷⁻¹⁹⁾. However, it is of interest to examine the effect on deformation of interstitial impurities and of the α - γ transformation.

In iron containing interstitial impurities, nitrogen or carbon, the flow stress increases (and the ductility decreases) as the temperature rises in the blue brittle temperature range 250-350°C^(11, 20). Subgrains do not form and the dislocations are evenly and densely distributed⁽¹¹⁾ as in sub-zero deformation⁽¹⁹⁾; although in highly purified Fe and Fe-4% Mn

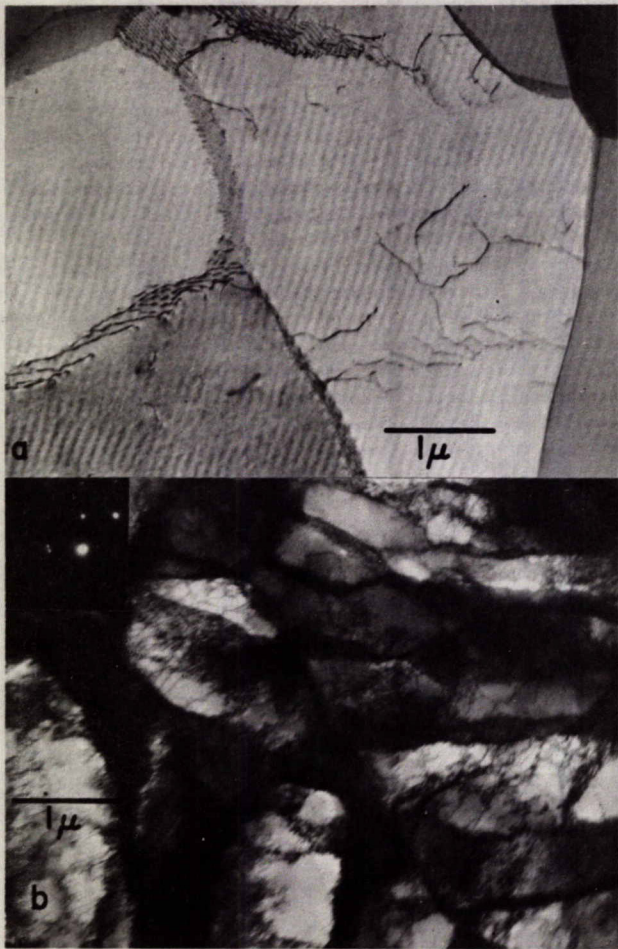


Figure 3 Cell size and perfection decreases as the stacking fault energy decreases in (a) aluminum (high purity) and (b) copper (OFHC) which have been hot rolled at 90% reduction in a single pass at a strain rate of 20 sec^{-1} and at 340° and 600°C (0.65 of their respective absolute melting points).

there is the normal formation of subgrains. Above and below this range the properties and structure of the interstitial-free and interstitial-rich iron are almost identical⁽¹¹⁾. The increase in flow stress in the blue brittle range is due to the drag exerted on moving dislocations by atmospheres of interstitial atoms which diffuse at approximately the same rate as the dislocations move^(11, 20). If the strain rate is increased, the temperature range for this effect must rise to match the velocities of the interstitial atoms and the dislocations. Increasing the carbon content up to 0.1% increases both the flow stress and the ductility⁽²¹⁾.

In the austenite range, for both iron and plain carbon steels, the flow stress decreases and the ductility increases as the temperature is raised^(13, 14, 21). Subgrains form and their size

increases with temperature of deformation^(14, 22). In the ausforming process, the dislocations introduced during the hot working are carried over into the martensite and increase its strength⁽²³⁾. Just above the transformation temperature, the flow stress of austenite is higher and its ductility lower than those of ferrite below the transformation temperature^(13, 14, 21). This situation is partially explained by the lower self-diffusivity in austenite, which decreases the rate of dynamic recovery (the ratios of the creep rates and of the diffusion coefficients of austenite and ferrite under the same conditions have been observed to be nearly equal⁽⁸⁾). The differences are, however, larger than can be attributed to inequality in diffusivity⁽¹⁴⁾; and probably reflect the greater difficulty of climb and cross-glide in the fcc austenite. With dissolution of more carbon (up to 1%), the ductility increases whereas the flow stress remains constant because carbon increases the self-diffusion rate⁽²¹⁾.

Just below the transformation temperature, the ductility of ferrite drops to the same level as that of austenite; this effect extends to lower temperatures for less pure irons^(13, 14, 21, 24). This loss of ductility is caused by formation of austenite at grain boundaries to which impurities have segregated^(14, 24). A similar decrease in ductility is observed in chromium steels, not only at the α - γ transformation but also at the γ - δ transformation⁽²⁴⁾.

PROPERTIES OF HOT-WORKED METALS

The room-temperature strength of hot-worked material is usually greater than that of recrystallized material^(12, 13, 15, 25-28). As the temperature is raised and the strain rate lowered, the strength decreases until it is little greater than that of recrystallized material⁽²⁶⁻²⁸⁾; this variation is consistent with the dependence of room-temperature strength on subgrain size ($\sigma \propto d^{-1/2}$)^(12, 27) and with the dependence of subgrain size on temperature and strain rate. By means of controlled hot working, a metal may be deformed directly to temper and would have the added advantage of greater ductility and lower internal stresses than cold-worked metal of the same strength. The polygonized substructure generally has improved creep resistance and in most cases exhibits good stability in service at elevated temperatures^(15, 22, 27-32).

Since the hot-worked structure usually has a preferred orientation and is stronger and less ductile than recrystallized metal, it is not as suitable for further forming operations; however, holding at the working temperature usually induces recrystallization. For metals which undergo limited dynamic recovery the time for recrystallization is so short that it is difficult to quench in the as-worked structure⁽⁹⁾; whereas for metals which undergo extensive recovery or contain impurities or precipitates, the time may be very long. In general the rate of recrystallization at a given forming temperature increases with strain rate because of the increased dislocation density. The rate of recrystallization also increases as the forming temperature rises because of the higher nucleation

rate even though the dynamic recovery during deformation was greater.

DYNAMIC RECRYSTALLIZATION

Sellars and Tegart^(33, 34) maintain that steady-state deformation in copper and other metals of low stacking fault energy is the result of dynamic recrystallization, that is, recrystallization during the course of deformation. Recrystallized grains form and grow at different points in the specimen. These regions then deform at a lower flow stress than those which have not yet recrystallized. As the earliest recrystallized regions harden, fresh recrystallized grains form elsewhere; by the time the recrystallization has consumed the original matrix, a second generation of recrystallized grains are appearing. It is conceded that if recrystallized grains nucleate their rate of growth will be quite rapid; what is difficult to envisage is the nucleation of the new grains.

Nucleation requires the formation of a region enclosed by a boundary of which all, or part, is of high misorientation ($> 10^\circ$)⁽³⁵⁾. This could be an existing high-angle boundary which migrates as a result of non-uniform strain; this is unlikely because the strain is uniform^(7, 35) and metallographic evidence points to true nucleation^(7, 36). It could be a cell wall of high density produced directly by deformation, however dynamic recovery and repolygonization during hot working would tend to prevent the formation of such boundaries. Adjacent cells could coalesce into a large subgrain as the dislocations from their shared boundaries migrated to their external boundaries to increase its misorientation⁽¹⁷⁾; again dynamic recovery and repolygonization would break down the external boundary and create new internal walls. The proponents of dynamic recrystallization have not proposed any alternate theory.

This dynamic recrystallization hypothesis was initially founded on the observation of recrystallized microstructures in quenched hot-worked specimens⁽³⁶⁾. It is now fairly certain that these structures formed during cooling since they appear undeformed and since recrystallization has been observed to occur in copper within a time much shorter than the quench time⁽⁷⁾. Other evidence supporting the dynamic recrystallization hypothesis has been shown to be inconclusive^(1, 7). If recrystallization did occur during deformation the new grains would be expected to contain different degrees of deformation; no evidence of this has as yet been presented.

Shapiro and Dieter⁽³⁷⁾ and Tegart⁽²⁾ have recently presented new evidence supporting dynamic recrystallization. For both copper and nickel, a minimum is observed in the ductility at the low end of the hot-working temperature range. At the minimum fracture is a result of the formation and extension of pores at the original grain boundaries. These pores are, presumably, the result of grain-boundary sliding which, although it is a very small fraction of the total strain at strain rates greater than 10^{-3}sec^{-1} ⁽³⁸⁾,

still occurs and can produce pores because of the large strain (≈ 2). At temperatures above those at which low ductility is observed, the ductility markedly increases although pores have nucleated early in the test. It was postulated that recrystallization occurs removing the old boundaries and preventing further extension of the crack. Termination of the test before fracture and rapid quenching do indeed reveal new grains around the cracks; however, the new grains have not yet been examined to determine if they are deformed. If these new grains did form during the deformation, it may be that relaxation around the crack reduced the flux of mobile dislocations to a level that did not hinder nucleation. If such a cluster of new grains formed, dynamic recrystallization could spread from it because the soft grains would deform more than average thus causing a lower strain rate in adjacent hardened regions in which nucleation could then occur.

If the recrystallized grains have actually formed only after deformation ceased, the cessation of pore growth may be explained by dynamic recovery. Aluminum which does not undergo dynamic recrystallization exhibits as great ductility as copper or nickel. In this case, the growth of pores is inhibited by migration of the boundary⁽³⁹⁻⁴¹⁾. Migration often occurs in conjunction with subgrain formation in the grains and results in a scalloped or corrugated boundary^(27, 28) which is resistant to sliding and hence to further pore formation. The ductility minimum is then a result of insufficient dynamic recovery and related boundary migration to stop propagation of the pores. Pore formation may be the cause of the hump in the stress-strain curves in torsion tests, a hump which is usually absent in compression tests.

Dynamic recrystallization may occur in multi-phase materials. Robbins, Shepard and Sherby⁽²¹⁾ have observed that pearlite rapidly spheroidizes during high-temperature deformation; while the evidence is not conclusive, it appears that recrystallization is involved. Packer and Sherby⁽⁴²⁾ propose that continual recrystallization is the mechanism in super-plastic deformation of eutectoid Al-Zn alloys. The deformation of an agglomerate of phases with different plastic properties is undoubtedly inhomogeneous and this is probably responsible for the recrystallization.

Dynamic recrystallization has been observed in creep^(43, 44) but in this case does not give steady-state deformation but rather grossly disturbs steady state by occurring in repeated waves with sudden, large increases in strain rate. The mechanism of recrystallization appears to be strain-induced bulging or migration of pre-existing high-angle boundaries^(9, 43, 44). Since this mechanism requires an inhomogeneity of strain which greatly diminishes as the strain exceeds 0.5, it is unlikely to be the mechanism in hot working where the strain for the postulated dynamic recrystallization exceeds 1^(9, 18). It possibly could occur in a situation where the strain is grossly inhomogeneous because of the mode of deformation, as it is from surface to centre in solid torsion specimens.

High-Temperature Flow Curves

When a metal is deformed at constant strain rate, $\dot{\epsilon}$, at a temperature, T , above half of its absolute melting point, the flow stress, σ , generally increases up to a strain between 1 and 2 and then remains constant for further strain which may be in excess of 10 (Fig. 4). Throughout this stage, known as steady-state deformation, there is no work hardening which is the result of the persistence of a constant dislocation density. Although this is considerably different from deformation at low temperatures where strain hardening continues up to fracture, it is consistent with the over-all pattern which shows a decrease in the rate of strain hardening beyond some large strain whose magnitude decreases with increase in deformation temperature.

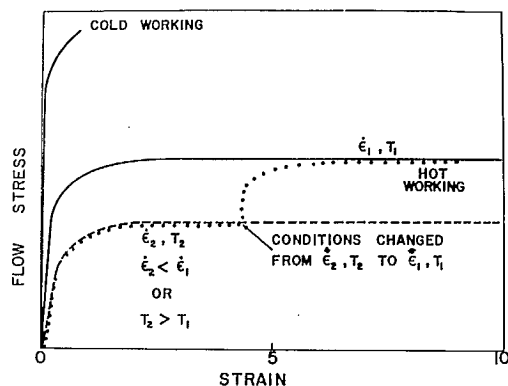


Figure 4 Flow curves in hot working exhibit horizontal, steady-state regions at high strains. The steady-state flow stress is characteristic of the forming conditions and changes to the new characteristic value if the conditions are altered.

The phenomenon of steady-state deformation also occurs in secondary creep and has been extensively studied. The controlled variables in creep are stress and temperature which determine the strain rate; in hot working, the controlled strain rate and temperature determine the flow stress. The primary decrease in creep rate to that of the steady state indicates a strain hardening equivalent to the rise in flow stress preceding steady-state hot working. Strain rates between 10^{-12} and 10^{-1} sec^{-1} are considered as creep, those between 10^{-1} and 10^4 as hot working and those above 10^4 as shock loading.

For a given alloy, the steady-state flow stress is generally independent of initial structure; however, in some cases, slow structural changes induced by the deformation delay the inception of steady state.⁽²¹⁾ Furthermore if the hot-working temperature or strain rate is changed during steady state, the flow stress changes gradually to the value characteristic of the new conditions (Fig. 4). In creep it is the rate of deformation that adjusts to changes in

temperature or stress⁽⁸⁾

Interdependence of Flow Parameters

Several empirical mathematical expressions have been employed to relate σ , $\dot{\epsilon}$ and T . The one which most closely conforms to the experimental data (Fig. 5a) and which actually is a synthesis of the others is

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

$$\text{or } A(\sinh \alpha\sigma)^n = \dot{\epsilon} \exp(\Delta H/RT) = Z \quad (2)$$

where A , α , n and ΔH are empirical constants, and R is the gas constant. Equation (1) may be considered the creep form⁽⁸⁾ and equation (2) the hot-working form; however, both forms (1) and (2) are used in hot working as was initially put forward by Sellars and Tegart⁽³³⁾. Jonas and Wong⁽⁴⁵⁾ have recently shown that, for aluminum, these equations apply to collected high-temperature-deformation data which span ten orders of magnitude of strain rate.

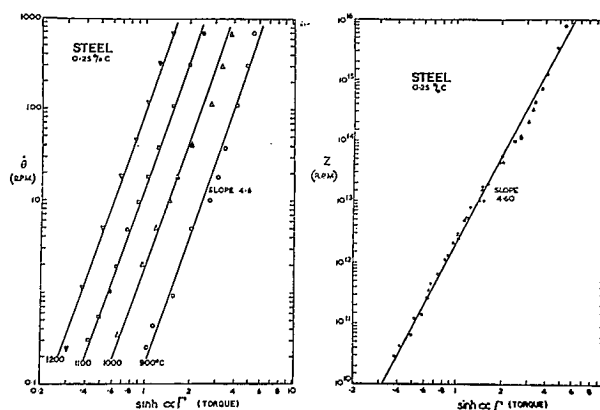


Figure 5 A log-log plot of the strain rates against the steady-state flow stresses results in a series of parallel straight lines, one for each temperature (a). The spacing between the lines can be used to determine the activation energy which is used in the Zener-Hollomon parameter (b). Sellars and Tegart⁽³³⁾

The dependence on temperature through an Arrhenius term, $\exp(-\Delta H/RT)$, indicates that the rate-controlling mechanisms of deformation are thermally activated. Thus ΔH is the height of the energy barrier blocking the atomic rearrangements involved in the mechanism (Fig. 6) and the exponential term expresses the availability of that quantity of energy at the temperature of deformation. The activation energy measured is that of the rate controlling mechanism although other mechanisms may be operating, which have lower activation energies and occur whenever required. For the hot working of metals, the activation energy is usually almost equal to that for self-diffusion and for creep. This equality leads one to postulate that the rate-controlling

mechanisms of hot working and of creep are the climb of either edge dislocations or edge jogs in screw dislocations. It is now seen that the hot-working temperature range is above half the absolute melting point because it is only above that temperature that the rate of vacancy migration is appreciable. In the cases where the activation energy observed in hot working is higher than that of self-diffusion, it does not seem to correspond to any known mechanism, although it has been interpreted as an indication of recrystallization in the course of deformation.

Combined Temperature-Strain Rate Parameter

If the Arrhenius term is coupled with the strain rate (Eqn. 2), the dependence of the flow stress on strain rate and temperature can be summarized in the Zener-Hollomon parameter⁽¹⁾, Z , which is normally maintained constant during a hot-working test. The results in Figure 5 show that deformations under different conditions which have the same Zener-Hollomon parameter, require the same flow stress. This is to be expected since, at a given value of Z , there are the same number of thermally activated events per unit strain because a decrease in strain rate compensates for the reduction in the number per unit time resulting from a lower temperature. From a graph such as Figure 5b the required operating temperature can be calculated for a desired forming rate and maximum force.

Stress-aided Activation

The hyperbolic-sine stress-function can be interpreted in the following way⁽⁴⁶⁾

$$\begin{aligned}
 & A(\sinh \alpha\sigma)^n \exp(-\Delta H'/kT) \\
 &= A/2(\sinh \alpha\sigma)^{n-1} \left[\exp(\alpha\sigma) - \exp(-\alpha\sigma) \right] \exp(-\Delta H'/kT) \\
 &= A/2(\sinh \alpha\sigma)^{n-1} \left[\exp\left(-\frac{\Delta H' - V\sigma}{kT}\right) - \right. \\
 & \quad \left. \exp\left(-\frac{\Delta H' + V\sigma}{kT}\right) \right] \quad (3)
 \end{aligned}$$

where $\Delta H' = \Delta H/6 \times 10^{23}$ is the activation energy per atom, k is Boltzman's constant and $V = \alpha kT$, is the activation volume. For aluminum⁽⁴⁷⁾ which at 500°C has an observed $\alpha = 4 \times 10^{-9}$ (dyne/cm²)⁻¹, and for copper⁽³³⁾ at 850°C with $\alpha = 3 \times 10^{-9}$ (dyne/cm²)⁻¹ the activation volumes are approximately $100b^3$ (b Burger's vector); which is roughly in agreement with those observed in creep⁽⁴⁸⁾ or in differential tension tests^(49, 50).

The activation volume is a measure of the effect of stress in aiding the dislocations to surmount the energy barriers (Fig. 6). Physically the activation volume may be thought of as the volume swept out by the dislocation in a single activated event. The activation volume observed is interpreted as the motion of a screw dislocation resulting from the climb of an edge jog upon removal of one vacancy⁽⁴⁹⁾; the

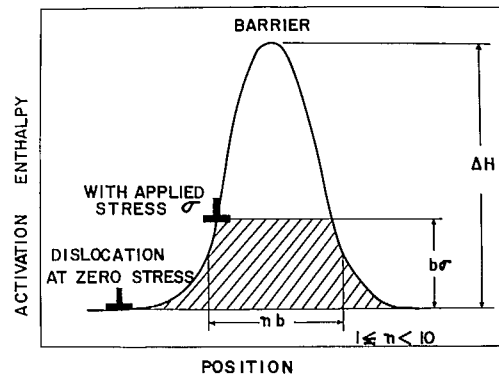


Figure 6 The activation energy required for a dislocation to surmount a barrier may be reduced by stress. The cross hatched area represents the work per unit length $nb \cdot b \cdot \sigma$; $b \cdot \sigma$ is the force per unit length.

dislocation of height b advances a distance b along a length $100b$ which is the average distance between the jogs (Fig. 7). In the model of Barrett and Nix⁽⁴⁹⁾, the screw dislocations are imagined to be moving at a rate determined by the jogs they are dragging; the term in the square bracket gives the effect of stress and temperature on the velocity of the dislocations.

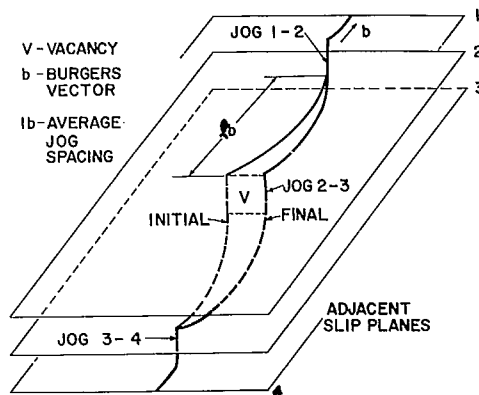


Figure 7 Displacement of a screw dislocation as a result of unit climb of a unit jog. With an average jog spacing lb , the work performed by stress σ is $lb \cdot b \cdot b \cdot \sigma$; lb^3 is known as the activation volume.

The remaining term $(\sinh \alpha\sigma)^{n-1}$ represents the effect of the dislocation substructure on the strain rate (from the creep viewpoint) or on the flow stress (hot-working viewpoint). In the Barrett and Nix model this would represent the variation in the mobile-dislocation density which has been observed to be proportional to σ^3 for creep. When extrapolated to the low stresses of creep, the hot working results essentially agree with this since with $n = 4.2$ as observed

for aluminum, iron and silicon steel⁽⁴⁷⁾ and for $\alpha\sigma < 0.7$

$$(\sinh \alpha\sigma)^{n-1} \approx (\alpha\sigma)^{3.2} \quad (4)$$

In hot working the cell diameter, d , is the only substructure parameter which has been measured and it has been observed to decrease as the flow stress (or Z) increases^(7, 9). Jonas et al⁽⁴⁷⁾ has shown that the diameter is related to the flow stress by the following equations:

$$d^{-1} = a + b\sigma \quad (5)$$

$$d^{-1} = p(\sinh \alpha\sigma)^q \quad (\text{Fig. 8}) \quad (6)$$

$$d^{-1} = r + s \log(\sinh \alpha\sigma) \quad (7)$$

The experimental values for the constants a , b , p , q , r and s are given for aluminum, iron and silicon steel⁽⁴⁷⁾. Eqn (5) is generally unsuited for extrapolation to creep because the computed subgrain diameter does not increase as rapidly as the experimental when the stress decreases; however, it has proved suitable for extrapolating to creep data on aluminum⁽⁷⁾. Eqns. (6) and (7) allow for a rapid increase in subgrain size at low stresses but there is insufficient creep data to test them.

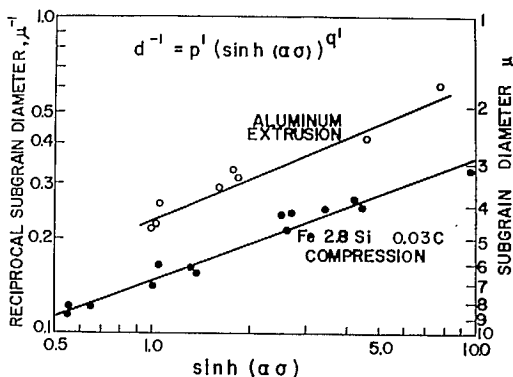


Figure 8 Relationship of subgrain diameter to flow stress in hot working. Jonas et al⁽⁴⁷⁾.

As the subgrain diameter d increases the dislocation density in the sub-boundaries decreases as the dislocations become more neatly arranged. As there is very little change in misorientation with subgrain size, the change must be taking place in the number of redundant dislocations in the boundaries. The variation in the number of dislocations per unit area of cell wall as a function of deformation conditions has not yet been measured; however, it is estimated that it varies inversely as cell diameter. If the number of dislocations per unit area of wall remains constant, the average dislocation density varies inversely as the cell size; thus, in the present case, the dislocation density varies inversely as the square of the cell diameter (d^{-2}). If the stress is

raised, the subgrain size decreases in proportion to $(\sinh \alpha\sigma)^{-0.4}$, since, in eqn. 6, q is approximately 0.4⁽⁴⁷⁾, and, therefore, the dislocation density varies as $(\sinh \alpha\sigma)^{0.8}$. The mobile dislocation density must increase as $(\sinh \alpha\sigma)^{3.2}$. Therefore, with increase in stress the mobile dislocation density increases approximately as the fourth power of the density of dislocations in the cell walls. No measurements of mobile dislocation densities have as yet been reported for hot-worked metals.

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