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Effect of Stacking-Fault Energy on Thermal Activation of Attractive Junctions in Change-in-Stress Creep of FCC Metals

Y. V. R. K. PRASAD, D. H. SASTRY, K. I. VASU
Materials Research Group, Metallurgy Department, Indian Institute of Science, Bangalore, India

Change-in-stress creep experiments involving increments as well as decrements of stress have been conducted on Ag, Au, Cu, Ni, Pb, Pd and Pt at temperatures at and below room temperature. The activation volumes (v) evaluated from the flow parameter, B (given by $B = (\Delta \ln \dot{\epsilon}/\Delta\tau) = v/kT$) obtained by making stress increments, were found to be in the range 10^2 to 10^4 $\mathbf{b}^3$, indicating the rate-controlling mechanism to be the intersection of glide and forest dislocations. The value of the flow parameter was found to be higher for creep following stress decrements than increments, the difference being a decreasing function of stress and temperature. The results were analysed in terms of the model, earlier suggested for aluminium, based on the formation of attractive junctions on stress decrement which causes a change in the number of elements participating in the activation event. The stacking-fault widths and energies evaluated for these metals based on the above model are found to be in good agreement with the reported values.

1. Introduction

It is now generally accepted that low-temperature creep in fcc metals is a process of strain-hardening and that its rate is controlled by a mechanism of intersection of glide and forest dislocations. On the basis of the model of intersection [1], the results of change-in-stress [2-4] and change-in-temperature [4-7] creep experiments were analysed to evaluate respectively the activation volumes and activation energies in various fcc metals. As regards the change-in-stress creep experiments, most of them involved stress increments and relatively little study has been made with stress decrements. Recent experiments conducted by us on some fcc metals, however, showed [8, 9] that the activation volume (or the corresponding flow parameter B) obtained after stress increments is considerably different from that obtained after stress decrements. It was also found that such behaviour was absent in cph metals [10-13]. Further, the results on aluminium [9] showed that the difference in the B-values obtained for stress increments and decrements was negligible at higher stresses and temperatures. Among the various possible mechanisms considered to explain the origin of this difference, it has been found [9] that a model based on the formation of attractive junctions on stress decrement, resulting in a change in the number of dislocation elements participating in the activation event, explains the effect satisfactorily.

The aim of the present investigation is to examine the mechanism by which the change occurs in the flow parameter on a stress decrement, considering the effect of stacking-fault energy in addition to the effect of stress and temperature. This will throw more light on the mechanism of low-temperature creep in fcc metals.

2. Experimental Procedure

The experimental data about the metals used in the present investigation are given in table I. All the samples were etched with appropriate chemical solutions before testing. Creep-testing was conducted on a constant-stress tensile creep unit described elsewhere [14]. In brief, it consists of a smooth calibrated contour-arm with the help of which the stress was maintained constant

during the test. The creep elongation was recorded as a function of time using a LVDT transducer and a millivolt recorder. The sensitivity of strain measurement was 10^{-5}. Tests were conducted at 87 and 197° K by surrounding the sample with liquid oxygen and solid carbon dioxide-alcohol mixture respectively. A stirred water bath was satisfactory at room temperature.

The testing procedure was as follows: the sample was initially loaded in appropriate steps until creep was observed on the recorder chart. When the creep rate slowed down, an incremental load was added and the creep curve recorded. The incremental load was then removed to record the decremental creep curve. This was followed by a major load addition and the small load addition/subtraction was repeated. With this procedure, it was possible to arrive at the change in the creep-rate following incremental as well as decremental stress at different stress levels and with different magnitudes of incremental/decremental loads.

3. Results

The variation of the creep-rate, $\dot{\epsilon}$, with strain was obtained by the method of graphical differentiation of the creep-strain, ϵ, versus log(time) plots at each of the major stresses. The change in the creep-rate associated with a stress increment or decrement was evaluated after extrapolating the log $\dot{\epsilon}$ versus ϵ plots (typical plots for Ag, Au, Pb, Pd and Pt are shown in fig. 1) to the value of strain at which the stress change has been made. The flow parameter B is defined [13] as:

$$B = \frac{\Delta \ln \dot{\epsilon}}{\Delta \tau} \simeq \frac{\delta \ln \dot{\epsilon}}{\delta \tau} = \frac{v}{kT} \qquad (1)$$

where ($\Delta \ln \dot{\epsilon}$) is the change in the creep-rate with the shear stress change $\Delta\tau$, v is the activation volume, k is Boltzmann's constant and T is the absolute temperature. The flow parameter was calculated at different stress levels for the stress increments as well as decrements. The value of shear stress τ for the above calculation was assumed to be half the tensile stress σ. From fig. 1, it can be seen that the values of Δ log $\dot{\epsilon}$ (and hence the flow parameter B) for decremental stress change are higher than for an equal incremental stress change for all the fcc metals tested. The stress-dependence of B obtained from creep-tests conducted at 87° K with different magnitudes of incremental and decremental stress for Ag, Au, Cu, Ni, Pb, Pd and Pt is shown in figs. 2a to 8a. These data show that the values of B corresponding to stress increments and their stress-dependence are both independent of the magnitude of the stress increment. On the other hand, the values of B corresponding to stress decrements are higher than those obtained for increments and vary with the magnitude of the decremental stress, the difference being dependent on the working stress. The higher the stress decrement, the larger the difference in the B-values. Generally, this difference between the B-values, hereafter denoted by β, decreases with increasing stress level and becomes negligible beyond a certain stress, depending on the magnitude of decremental stress and temperature; the effect of temperature

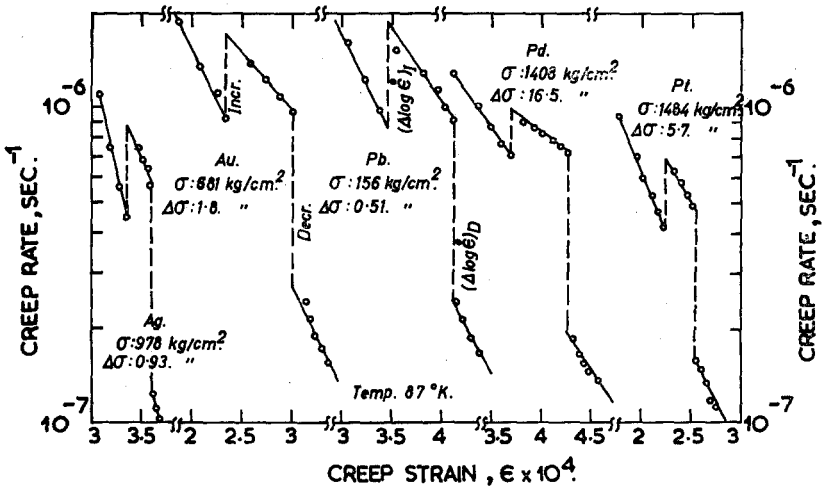


Figure 1 Typical variation of creep rate ($\dot{\epsilon}$) with creep strain (ϵ) during change-in-stress creep experiments at 87°. K on fcc metals: Ag, Au, Pb, Pd and Pt.

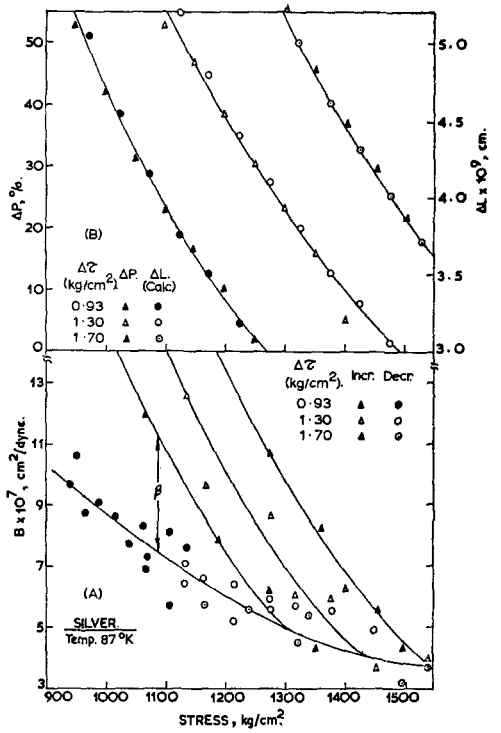
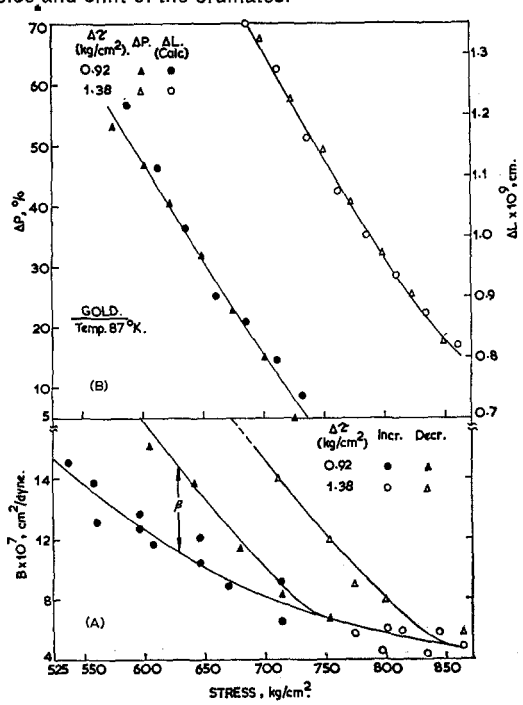


Figure 2 (A) Dependence of flow parameter B on stress for different magnitudes of incremental/decremental stress change for silver deformed at 87° K. (B) Change in the frequency factor ($\triangle P$,%) associated with different magnitudes of stress decrement, as a function of stress in silver. The calculated change in the junction length (ΔL) with stress is also shown, superimposed, by suitable choice and shift of the ordinates.



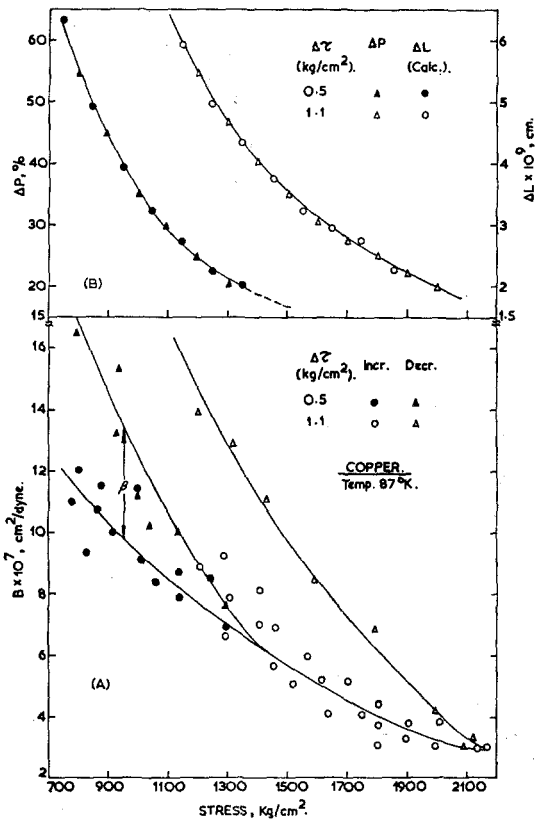


Figure 4 (A) Dependence of flow parameter B on stress for different magnitudes of incremental/decremental stress change for copper deformed at 87° K. (B) Change in the frequency factor ($\triangle P$,%) associated with different magnitudes of stress decrement, as a function of stress in copper. The calculated change in the junction length (ΔL) with stress is also shown, superimposed, by suitable choice and shift of the ordinates.

is clearly presented in figs. 9a to f. A comparison of figs. 2a to 8a with figs. 9a to f shows that β is smaller at higher temperature and is almost negligible at room temperature for Cu, Au, Ag and Pt. This is in agreement with the room temperature data reported by Wyatt [15] for Cu and by Thornton and Hirsch [2] for Au.

Figure 3 (A) Dependence of flow parameter B on stress for different magnitudes of incremental/decremental stress change for gold deformed at 87° K. (B) Change in the frequency factor (ΔP, %) associated with different magnitudes of stress decrement, as a function of stress in gold. The calculated change in the junction length (ΔL) with stress is also shown, superimposed, by suitable choice and shift of the ordinates.

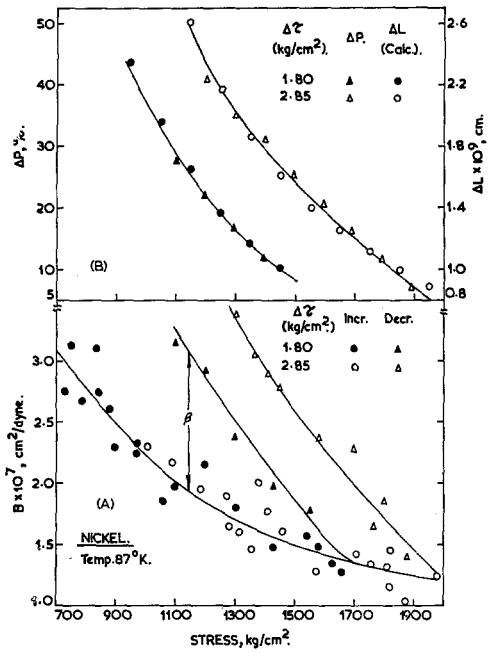


Figure 5 (A) Dependence of flow parameter B on stress for different magnitudes of incremental/decremental stress change for nickel, deformed at 87° K. (B) Change in the frequency factor (ΔP, %) associated with different magnitudes of stress decrement, as a function of stress in nickel. The calculated change in the junction length (ΔL) with stress is also shown, superimposed, by suitable choice and shift of the ordinates.

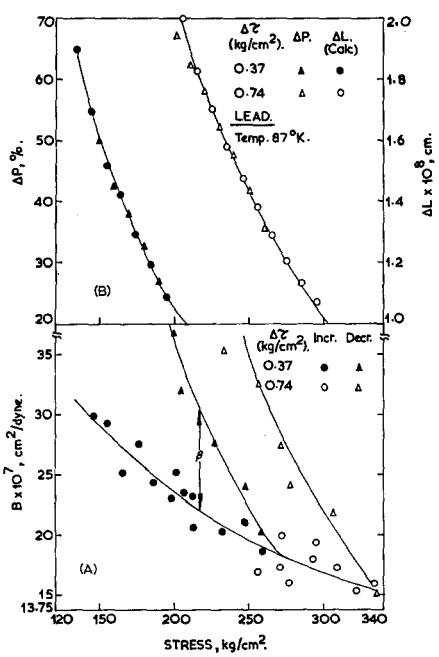


Figure 6 (A) Dependence of flow parameter B on stress for different magnitudes of incremental/decremental stress change for lead deformed at 87° K. (B) Change in the frequency factor (ΔP, %) associated with different magnitudes of stress decrement, as a function of stress in lead. The calculated change in the junction length (ΔL) with stress is also shown, superimposed, by suitable choice and shift of the ordinates.

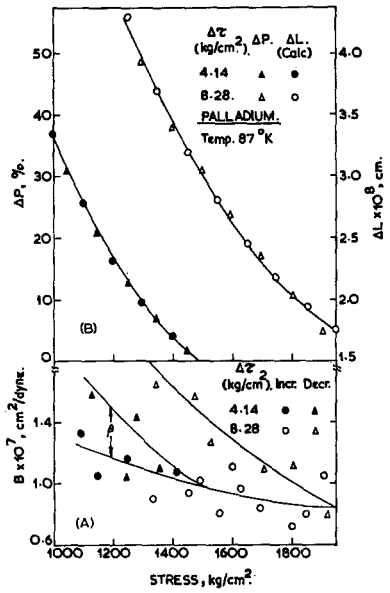


Figure 7 (A) Dependence of flow parameter B on stress for different magnitudes of incremental/decremental stress change for palladium deformed at 87° K. (B) Change in the frequency factor (ΔP, %) associated with different magnitudes of stress decrement, as a function of stress in palladium. The calculated change in the junction length (ΔL) with stress is also shown, superimposed, by suitable choice and shift of the ordinates.

4. Discussion

The range of the values of activation volume calculated from the flow parameter (corresponding to stress increments) using equation 1 for all the metals in the stress interval studied, is given in table II. The values fall in the range 10^2 to $10^4\mathbf{b}^3$ which is typical for the intersection mechanism [16]. In view of this and the earlier investigations [1-4, 17-20] it is reasonable to believe that the mechanism of intersection of glide and forest dislocations controls the rate of low-temperature creep. The creep-rate can then be represented [21] by:

$$\dot{\epsilon} = P \exp\{-[U_0 - v(\tau - \tau_G)]/kT\} \quad (2)$$

where P $(=NA\mathbf{b}v)$ is the frequency factor involving the number of dislocation elements per

TABLE I Experimental data on the metals used

Metal	Purity %	Specimen shape and size*	Treatment	Average grain dia. mm	Total no. of samples used
Ag	99.99	sheet, 0.58 mm thick 4.25 mm wide	annealed at 700° C for 2 h in air	0.005	6
Au	99.9	wire, 0.66 mm dia.	annealed at 750° C for 2½ h in air	~0.005	5
Cu	99.98	sheet, 1 mm thick 6 mm wide	annealed at 800° C for 3½ h in vacuum†	0.02	15
Ni	99.99	sheet, 0.58 mm thick 4.25 mm wide	annealed at 1000° C for 3 h in vacuum†	0.025	12
Pb	99.99	sheet, 2.1 mm thick 8.1 mm wide	annealed in oil bath at 100° C for 1 h	0.01	6
Pd	99.9	wire, 0.25 mm dia.	annealed at 1000° C for 4 h in vacuum†	<0.005	5
Pt	99.9	wire, 0.26 mm dia.	annealed at 1200° C for 4 h in air	<0.005	5

*The gauge length was 44.5 mm.
†The vacuum used was 2×10^{-6} torr.

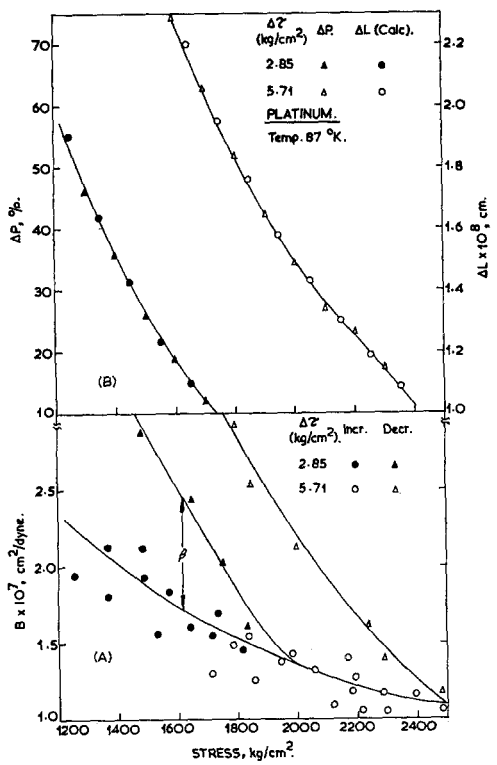


Figure 8 (A) Dependence of flow parameter B on stress for different magnitudes of incremental/decremental stress change for platinum deformed at 87° K. (B) Change in the frequency factor (ΔP, %) associated with different magnitudes of stress decrement, as a function of stress in platinum. The calculated change in the junction length (ΔL) with stress is also shown, superimposed, by suitable choice and shift of the ordinates.

cm³ participating in the activation event, the area swept per successful activation, the Burgers vector and the frequency of vibration of the dislocation line; U_0 is the total intersection energy, and $(\tau - \tau_G)$ is the thermal component of the flow stress. On differentiation of equation 2 with respect to τ at constant temperature, we get the flow parameter:

$$B = \frac{\Delta \ln \dot{\epsilon}}{\Delta \tau} = \frac{v}{kT} + \frac{\delta \ln P}{\delta \tau} - \frac{\delta U_0}{\delta \tau}. \quad (3)$$

This equation implies that there occurs a change in the frequency factor P as well as U_0 associated with a stress change, $\Delta\tau$. Earlier calculations of activation volume [2-4, 10-12] were based on the assumption that P and U_0 vary little with the magnitude of the stress change so that equation 3 reduces itself to equation 1. This has been found to be the case with stress increments in aluminium [9] and both stress increments and decrements in cph metals [10-13]. In view of the non-dependence of B on the magnitude of increment, and the similarity in the mechanism of deformation on increment between aluminium and the other fcc metals presently studied, it is reasonable to assume that the variation in P and U_0 will be negligibly small on stress increments.

The B-values for stress decrements during creep in all the fcc metals studied at 87° K (figs. 1, 2a to 8a) are, however, higher than those for stress increments; this would imply that some of the processes that occur during incremental creep do not do so on a stress decrement or that

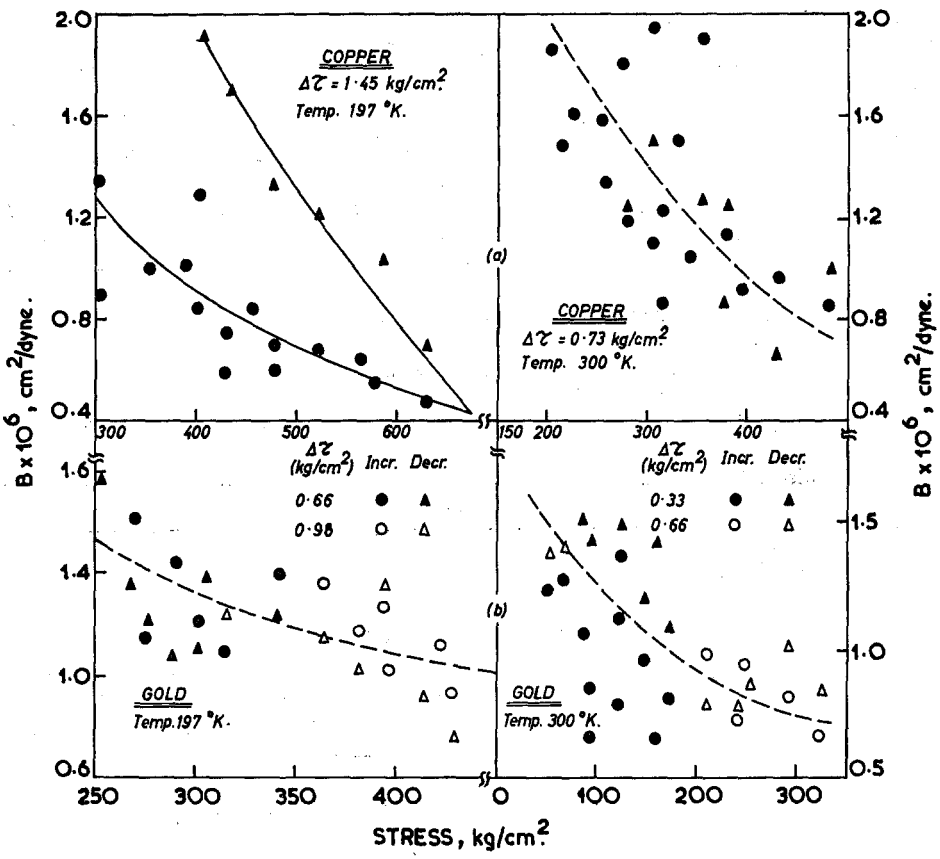
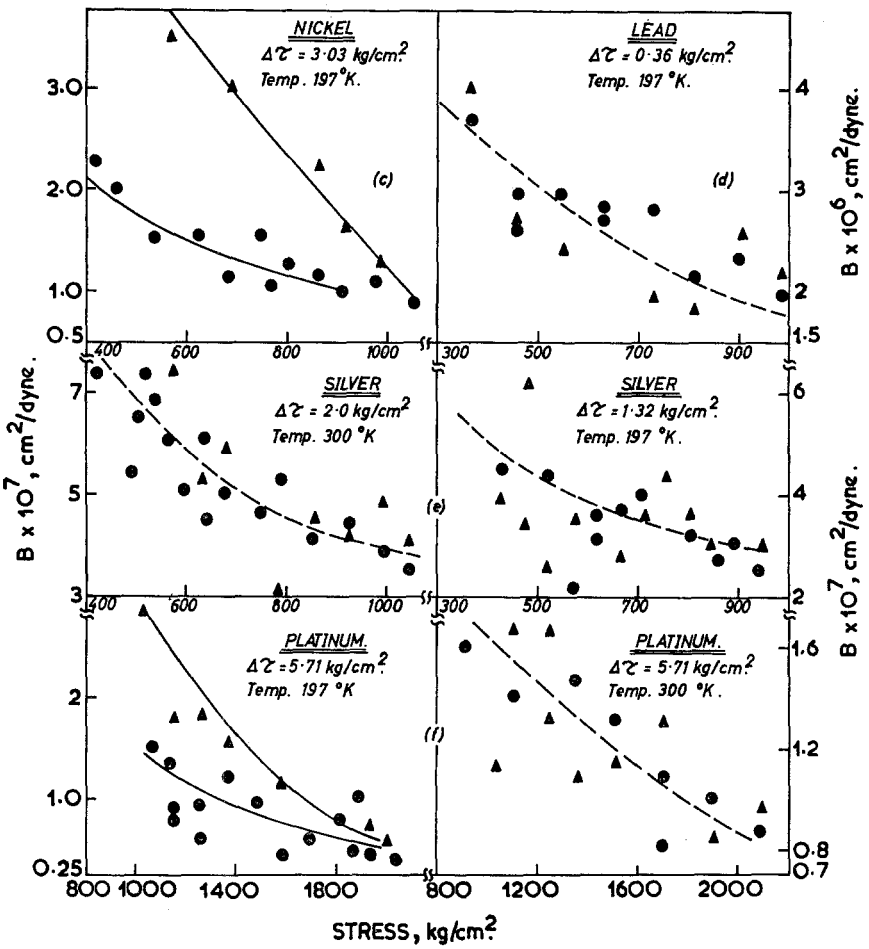


Figure 9 (a, b) Dependence of the flow parameter B on stress for copper and gold deformed at 197° K and 300° K. (c–f) Dependence of the flow parameter B on stress for nickel, lead, silver and platinum. The temperature of testing is indicated in the figure.

some additional dislocation interactions take place on stress decrement. If this were the case, the assumptions that P and U_0 vary little with the stress change, in such a way that equation 3 reduces itself to equation 1, may not be valid. Since the change in U_0 with stress, which is interpreted in terms of local atomic relaxations near the jogs [22], will only be small, we attribute [8] the higher values of B corresponding to stress decrements to a change in the frequency factor P upon a stress decrement. The fact that this change is not present in cph metals makes us think that processes which are not common in cph metals but are possible in fcc metals may be responsible for this change in P. Two such mechanisms which were considered earlier to explain the observations on aluminium [9] are: (i) cross-slip and (ii) attractive junction formation. The observed influence of stress, magnitude of the decremental stress, temperature and stacking-fault energy on β would enable us to decide which of the above two mechanisms is the actual cause for the change in the frequency factor.

4.1. Cross-slip

In view of the large number of slip systems in fcc metals, cross-slip occurs frequently. However, the range of activation volume (table II) observed in this investigation clearly shows that cross-slip is not the rate-controlling mechanism, since cross-slip requires that the activation volume be low (10^1 to $10^2\mathbf{b}^3$) [16]. Moreover, logarithmic creep is observed in all the metals at low temperatures, which implies the absence of any dynamic recovery involving cross-slip. Although cross-slip is not the rate-controlling mechanism, it might cause a change in the number of elements participating in the activation event. When a stress decrement is made during creep, there is an increase in thermal activation energy, proportional to the activation stress [1] (the difference between the "yield stress" and the applied stress). Consequently, the



dislocation segment which is held up at an obstacle and finds it difficult to penetrate owing to the larger thermal activation energy, can bypass it by cross-slip. This can cause a change in the number of elements participating in the intersection process.

Another type of cross-slip, suggested by McLean [23] to be intimately connected with intersection, is the process of "cutting". This occurs when two dislocations having equal and opposite Burgers vectors, while moving on intersecting planes, meet and annihilate part of their length; the remaining lengths will change their slip planes. Such a process would be favoured on a stress increment to a greater extent than on a stress decrement and hence can cause a difference between the number of elements participating in the activation event on a stress increment and decrement. Either of the types of cross-slip would help the cutting intersection to occur more frequently since the dislocation meets many more dislocations on the intersecting slip planes.

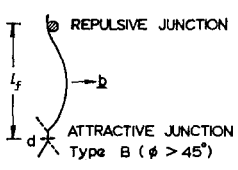


Figure 10 Schematic representation of a dislocation line held up at an attractive junction ($\phi > 45°$) and a repulsive junction during decremental stress creep.

If cross-slipping on a stress increment is responsible for the change in the number of elements participating in the activation event, the B-values obtained on a stress-increment should be dependent on the magnitude of the incremental stress. Since this is not observed, the process of cross-slipping on a stress increment cannot explain the present results. If cross-slip

TABLE II The stress ranges used and the activation volumes observed in fcc metals

Metal	Temperature (° K)	Stress range (kg cm^{-2})	Range of activation volume ($v \times 10^{21}$ cm^3)
Ag	87	900–1550	3.6– 13.2
	197	375– 805	5.4– 13.6
	300	390–1070	12.4– 31.0
Al*	87	200– 600	4.8– 14.4
	197	200– 350	30.0– 43.0
Au	87	525– 875	4.8– 18.0
	197	250– 450	21.7– 43.5
	300	50– 350	20.7– 62.1
Cu	87	750–2100	3.6– 14.4
	197	300– 650	10.8– 38.0
	300	175– 500	24.0– 74.5
Ni	87	700–2000	1.2– 3.7
	197	450–1000	2.2– 6.8
Pb	87	140– 400	18.8– 36.0
	197	36– 100	41.0–108.0
Pd	87	1100–1950	0.8– 1.7
Pt	87	1250–2500	1.5– 2.6
	197	1020–2050	0.6– 4.1
	300	800–1700	3.3– 7.0

*Calculated from the data given in [9].

is responsible for the change in B on decrement, there should be an increase in the number of elements participating in the activation process, and so the values of B on a stress decrement should be smaller than those for a corresponding stress increment. However, the experimental data clearly show the reverse. Further, at higher stresses and temperatures, the more frequent is the cross-slip and so, the difference should be higher. The experimental results again show a reverse trend. Moreover, the stacking fault energy has a profound influence on the occurrence of cross-slip. This has been considered by Seeger [21] and Pavlov *et al* [24]. An extended dislocation will have to be constricted before cross-slip and so requires higher energy than a unit dislocation. Thus in low stacking-fault energy metals (with wider stacking-faults), the energy required for the cross-slip is high and so the frequency of its occurrence would be low. If cross-slip is the mechanism causing the observed change, the difference β should not be observed in low stacking-fault energy metals like Ag, Cu, Au, Pb etc [24, 25] or, if it is observed at all, it should be to a very small extent. This is not in accordance with the present results since for all the metals, the flow parameter B was considerably higher on a stress decrement than on an increment (figs. 1, 2a to 8a). Taking all these arguments together, the present observations cannot be explained by the model based on cross-slip.

4.2. Attractive Junction Formation

Because of the strong elastic interactions between glide and forest dislocations in an fcc lattice, the formation of attractive junctions is favoured [26]. By detailed calculations Saada [27] and McLean [23] have shown that the formation of attractive junctions plays an important role in strain-hardening. The attractive junctions were found to be of two types, depending on whether or not the jog-formation at these junctions needs thermal fluctuations to augment the applied stress. For angles of meeting $\phi < 45°$ (type A), the stress required to break the junction reaches a peak before the nodes come together and then the jog-formation takes place. In this case the jog-formation occurs unaided by thermal fluctuations. However, for dislocations meeting at $\phi > 45°$ (type B), the resistance due to jog-formation adds to the junction resistance and

the jogs form only when thermal fluctuations augment the applied stress. The length of a type B junction dislocation at which a jog is about to form will, on stress decrement, increase, due to the relaxation of the dislocation network. Inasmuch as the energy provided by thermal fluctuations will be insufficient to break the increased length of the junction, thermally activated jog-production cannot occur at these places; this would reduce the number of elements participating in the activation event and thus bring about a change in the frequency factor. In the following is discussed the effect of stress, magnitude of the stress decrement, temperature and stacking-fault energy on β, on the basis of the model of attractive junction-formation.

4.2.1. *Effect of Stress and Magnitude of Stress Decrement*

The effect of stress on β is due to a change in the length of the junction with stress, on a stress decrement. At higher stresses, the length of the attractive junction that can form would be smaller, which would result in a higher probability of thermally activated jog-production at these junctions. Hence the change in the number of elements participating in the activation event would be smaller on stress decrement. One would then expect only lower β-values at higher stress levels. This is in good agreement with the observed trend of the variation of β with stress in all the metals (figs. 2a to 8a).

The change in the junction length, ΔL, on a stress decrement could be computed as a function of stress using the equation for the stress required to break the attractive junction, as given by Saada [27]:

$$\tau_G = \frac{\alpha G \mathbf{b}}{\ell_f} \qquad (4a)$$

where α is a constant with a value between 0.2 and 0.4, G is the shear modulus and ℓ_f is the average distance between the junctions as decided by the forest spacing. When a stress decrement, $\Delta\tau$, is made, there would be an increase in the junction length proportional to $\Delta\ell_f$, so that

$$\tau_G - \Delta\tau_G = \frac{\alpha G \mathbf{b}}{\ell_f + \Delta\ell_f}. \qquad (4b)$$

By assuming the junction length, L, to be one half of ℓ_f, and with the help of equations 4a and 4b, the change in the junction length ($\Delta L = \Delta\ell_f/2$) on decrement is evaluated as a function of stress and plotted in figs. 2b to 8b for the decremental stress values used for different metals. However, equations 4a and b have to be used with caution for the low stacking-fault energy metals since they are derived on the assumption that the dislocations are unsplit. As it is believed [27] that the effect of split dislocations may only be to increase the stress level, leaving the relation almost unchanged, we make the above assumption and evaluate ΔL-values. From the experimentally observed stress-dependence of β at 87° K (figs. 2a to 8a) and with the help of equation 3, the percentage change in the frequency factor, ΔP, %, associated with a stress decrement is evaluated and also plotted in figs. 2b to 8b as a function of stress. In this calculation, the effect of stress on U_0 is neglected. By a process of trial and error, the ordinate scales are chosen and shifted so as to optimise the agreement between the stress-dependence of the two variables. It is to be noted that the stress-dependence of ΔL and ΔP, %, agreed very well for the decremental stress values used, for all the metals.

4.2.2. *Effect of Temperature*

According to the model based on the formation of attractive junctions on decrement, the effect of temperature on β would be through the amount of energy that could be supplied by thermal fluctuations. Since the thermal energy supplied would be larger at higher temperature, the probability of the junction of a given length getting activated would be higher, and the change in P would be smaller. This would result in the participation of a larger number of attractive junctions in the activation event after decrement. The difference in B, associated with decrements or increments, would therefore be smaller at higher temperatures, in conformity with the present results for all the metals (figs. 9a-f).

4.2.3. *Effect of Stacking-Fault Energy*

The effect of stacking-fault energy (SFE) on the model based on attractive junction-formation is that in low SFE metals, as compared with high SFE metals, a smaller length of junction would be enough to cause an equal change in P, since in addition to the line tension, the constriction energy (proportional to stacking-fault width) is also provided for breaking the junction. Although the dislocation dissociations at triple nodes in a two-dimensional network have been analysed in detail [28,29], not much work seems to have been done regarding a three-dimensional

TABLE III Values of $v_{J\tau}$, the stacking fault widths and energies for the fcc metals

Metal	$v_{J\tau} \times 10^{11}$ (ergs)	$(\tau_0/\tau_G\dagger)$	$G \times 10^{-11}$ (dynes cm^{-2})†	Stacking-fault width, d (Å)	Most probable values of stacking-fault energy, γ (ergs cm^{-2})
Ag	2.9	1.2‡	2.9	37	20 ± 5
Al*	0.3	1.4	2.7	3.5	170 ± 34
Au	3.4	1.35	2.8	37	17 ± 8
Cu	2.4	1.2	4.6	22	35 ± 10
Ni	0.7	1.5	7.8	3	390 ± 110
Pb	1.5	1.2‡	0.75	220	11 ± 5
Pd	0.4	1.4‡	4.9§	2.3	420 ± 120
Pt	1.4	1.35‡	6.2	7.5	167 ± 75

*Obtained from data given in [9].
†Data quoted by Thornton and Hirsch [2].
‡Assumed.
§Quoted from Pavlov *et al* [24].

network. Some of the dissociations, however, were analysed by Lomer and Cottrell [30] and Hirth [31]. The junction dislocation, being on the line of intersection of the two {1 1 1} planes, may dissociate, for example, according to the reaction:

$$\underline{DA} \rightarrow \underline{D\alpha} + \underline{\alpha\delta} + \underline{\delta A}\,.$$

The partial dislocations $\underline{D\alpha}$ and $\underline{\delta A}$ repel each other and move away from the line of intersection, leaving a sheet of stacking-fault joining them to the stair-rod dislocation $\underline{\alpha\delta}$. For this dissociated junction dislocation to break, the nodes will have to come together. This requires the constriction of the extended junction dislocation. As the constriction requires higher energy, a smaller increase in the length of the junction would be enough to bring about an equal change in P as in a high SFE metal: comparison of the ΔL and ΔP scales in figs. 2b to 8b show that this is so.

The most probable limits of SFE could be evaluated from the experimentally observed β-values, using a model illustrated in fig. 10. According to this model, the creep rate on decrement, $\dot{\epsilon}^D$ is controlled by the mechanism of overcoming the attractive junctions of type B. This mechanism is assumed to involve a simple process of mechanical constriction of the extended junction dislocation as the nodes come together. The creep-rate following a stress decrement, $\dot{\epsilon}^D$, can be represented as:

$$\dot{\epsilon}^D = \dot{\epsilon} \exp\{-[E - (\tau - \tau_G)\,\mathbf{b}d.\ell_f]/kT\} \quad (5)$$

where $\dot{\epsilon}$ is the creep-rate prior to decrement, E is the total energy required to break the junction dislocation, $(\tau - \tau_G)$ is the thermal component of flow stress, τ_G is the athermal flow stress controlled by the breaking of attractive junctions of type A, d is the width of the barrier in the direction of the Burgers vector $\mathbf{b}$ and ℓ_f is the distance between the junctions. In formulating equation 5, a "square-topped" barrier has been assumed for the sake of mathematical simplicity. If the magnitude of the decrement is changed from $\Delta\tau_1$ to $\Delta\tau_2$, we get

$$kT \ln(\dot{\epsilon}_2{}^D/\dot{\epsilon}_1{}^D) = (\Delta\tau_1 - \Delta\tau_2)\,\mathbf{b}.d.\ell_f \quad (6)$$

where $\dot{\epsilon}_1{}^D$ and $\dot{\epsilon}_2{}^D$ represent the creep-rates after stress decrements $\Delta\tau_1$ and $\Delta\tau_2$ respectively. In arriving at equation 6, the change in E and ℓ_f with the change in the magnitude of stress decrement is neglected. Inasmuch as the change in the creep-rates after the two decrements is due to a change in P alone, we can write equation 6 as:

$$kT \ln(P_1{}^D/P_2{}^D) = (\Delta\tau_1 - \Delta\tau_2)\,\mathbf{b}.d.\ell_f \quad (7)$$

or

$$kT(\Delta \ln P)\frac{\tau}{\Delta\tau_1 - \Delta\tau_2} = \mathbf{b}.d.\ell_f.\tau\,. \quad (8)$$

Replacing the ratio of $(\tau/\Delta\tau)$ by the ratio of their tensile counterparts $(\sigma/\Delta\sigma)$, equation 8 can be written as

$$v_{J\tau} = kT(\Delta \ln P)\,[\sigma/(\Delta\sigma_1 - \Delta\sigma_2)] \quad (9)$$

where the junction activation volume v_J, is given by $\mathbf{b}.d.\ell_f$. The average values of $v_{J\tau}$ calculated for various metals tested at 87° K are given in table III. From equations 9 and 4a, ℓ_f can be eliminated with a view to getting the barrier width d, provided $v_{J\tau_G}$ values are obtained. The values of $v_{J\tau_G}$ can be calculated from $v_{J\tau}$ by following the procedure adopted by Thornton

and Hirsch [2], using the data on the temperature-dependence of flow stress. However, this procedure has a limitation in that it assumes that the mechanism controlling the initial flow stress also controls the subsequent strain-hardening, which, according to the present understanding, is not too far from being true. Making the above assumption, we calculate $v_{J\tau_G}$ from $v_{J\tau}$ by dividing the latter by the ratio (τ_0/τ_G), where τ_0 is the extrapolated value of the flow stress at 0° K. The values of (τ_0/τ_G) quoted by Thornton and Hirsch [2] are used in the calculation and are tabulated in table III. From equations 9 and 4a, it can be shown that

$$d = \frac{kT(\Delta \ln P)\,[\sigma/(\Delta\sigma_1 - \Delta\sigma_2)]}{(\tau_0/\tau_G)\,\alpha\, G\, \mathbf{b}^2} \,. \quad (10)$$

Since the width of the barrier, taken in the direction of the Burgers vector **b**, is equal to the stacking-fault width, equation 10 gives directly the stacking-fault widths from the experimentally measured parameters $(\Delta \ln P)$ and $[\sigma/(\Delta\sigma_1 - \Delta\sigma_2)]$ for various metals. The calculated values of stacking-fault widths are given in table III. Taking these as the equilibrium widths, the stacking fault energy γ is calculated from the equation:

$$\gamma = \frac{G\, a^2}{24\pi d} \quad (11)$$

where a is the lattice parameter. The range of the values obtained for the SFE is indicated in table III. The values are in good agreement with those reported in the literature as obtained from other methods [24, 25], thus giving further support to the model based on the formation of attractive junctions on a stress decrement.

The above discussion shows that, on stress decrement, there occurs a change in the number of elements participating in the activation event identified as the intersection of glide and forest dislocations; this change is caused by the formation of an attractive junction dislocation on stress decrement. In view of this, the values of B corresponding to stress increments alone are to be taken for the calculation of activation volume using equation 1, since on a stress decrement a change occurs in the frequency factor P. However, if such a change does not occur on a stress decrement, as in cph metals, the values corresponding to stress decrements would be more accurate compared with those obtained with stress increments.

5. Conclusions

(i) The values of the activation volume obtained by making stress increments during low-temperature creep of fcc metals fall in the range 10^2 to $10^4\mathbf{b}^3$, thus indicating that the rate-controlling mechanism is the intersection of glide and forest dislocations.

(ii) The activation volume values obtained by making stress decrements are larger than those obtained on stress increments and are dependent on the magnitude of decremental stress; this is attributed to the change in the number of elements participating in the activation event.

(iii) Among the models considered, mainly based on cross-slip and attractive junction-formation on decrement, to explain the change in the number of elements participating in the activation event, the experimental results could be satisfactorily explained on the basis of the formation of attractive junctions on decrement. Thus, in addition to the thermal activation of repulsive junctions during creep at low temperatures, thermally activated jog-production also occurs at some of the attractive junctions.

(iv) The stacking-fault energy values evaluated from the decremental creep data and based on the model of attractive junction formation, are in good agreement with those reported for Ag, Al, Au, Cu, Ni, Pb, Pd and Pt as obtained by other methods.

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