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Plastic flow in bcc solid solutions

M. Z. BUTT, K. M. CHAUDHARY
Department of Physics, Government College, Lahore, Pakistan

P. FELTHAM
Department of Physics, Brunel University, Uxbridge, UK

Available data on the temperature dependence of the CRSS of potassium, sodium, niobium, tantalum and of several binary bcc alloys are examined with reference to the kink-pair process of dislocation movement described by Butt and Feltham [1]. The model is found to account for the observations on the pure metals; equally the temperature and concentration dependence of the CRSS of the alloys are encompassed by it when the alloy-atom pinning of edge dislocations becomes strong enough to render the latter less mobile under stress than the screws.

1. Introduction

The possibility that in relatively concentrated bcc solid solutions the mobility of edge dislocations may be lower than that of the screws, and hence determine the critical resolved shear stress (CRSS), was recently considered by Butt and Feltham [2]. They confirmed that the kink-pair model of solid-solution hardening, initially developed for alloys with low Peierls potentials [1, 3, 4], which would be expected to hold if the movement of edge dislocations were rate determining, accounted quite well for the temperature and concentration dependence of the CRSS, τ, of single crystals of Nb–Mo and Nb–Re alloys containing 5 to 16 at% Mo and 5 to 9 at% Re, with

$$\tau = \tau_0\theta/[1+(1+\theta)^{1/2}]^2, \qquad (1)$$

where $\tau_0 = 4Uc^{1/2}/nb^3$, and $\theta = 4n^2Gb^3Uc^{1/2}/(mkT)^2$, m being generally equal to about 25. Here $\tau \rightarrow \tau_0$ as $T \rightarrow 0\,\mathrm{K}$, c is the alloy concentration expressed as atomic fraction, U the mean binding energy between an alloy atom and the dislocation considered, n the average number of interatomic spacings by which a dislocation bordered by a kink pair would advance in an activated jump, G is the shear modulus and b is the interatomic distance in the slip direction.

As kink-pair formation is also considered to facilitate the passage of dislocations over Peierls barriers, an equation of the same form as Equation 1 might also be expected to account for the temperature dependence of the CRSS of pure bcc metals. In this case U would be interpreted as the Peierls energy per interatomic spacing along the dislocation, c in Equation 1 being taken equal to 1 [5], and n would be anticipated to lie in the range $\frac{1}{2}$ to 2.

The main object of the present work was to examine available data on the temperature dependence of the CRSS of single crystals of relatively concentrated bcc binary alloys, as well as of bcc metals and, in the former case, also the influence of alloy concentration on the magnitude of τ, in the light of the model represented by Equation 1 and the associated expression for the activation volume

$$v = \tfrac{1}{2}v_0[(\tau/\tau_0)^{-1/2} + (\tau/\tau_0)^{-3/2}],$$
$$v_0 = \tfrac{1}{4}b^3n^2(Gb^3/Uc^{1/2})^{1/2}. \qquad (2)$$

A further relation, derivable from Equation 1, readily amenable to experimental study, refers to the ratio of the CRSS at two temperatures, $T_2 > T_1$. On writing $\theta = A^2/T^2$ for a specific alloy, and $\Delta\tau \equiv \tau_1 - \tau_2$, it is given by

$$\frac{\Delta\tau}{\tau_1} = 1 - \left[\frac{T_1 + (T_1^2 + A^2)^{1/2}}{T_2 + (T_2^2 + A^2)^{1/2}}\right]^2, \qquad (3)$$

and with typical values of A the ratio $\Delta\tau/\tau_1$ is usually not sensitively dependent on the concentration c.

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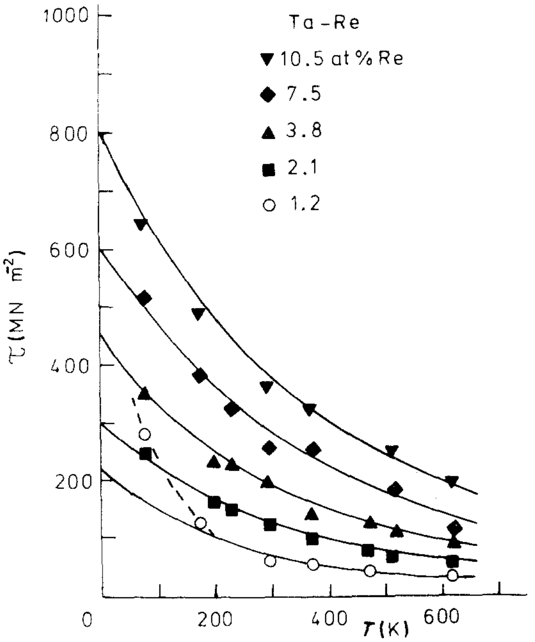

Figure 1 The CRSS of Ta–Re crystals. The curves were obtained by means of Equation 1 with parameters given in Table I; the experimental points were taken from Mitchell and Raffo [6].

2. Typical binary solid solutions

Figs. 1 to 5 show the correlation referred to in the case of relatively concentrated Ta–Re, Ta–Mo and Ta–W alloys; similar data for Nb–Mo are given in Figs. 6 to 8. The full lines drawn through the data points in the diagrams referring to the $\tau(T, c)$ relation represent Equation 1 with the parameters listed in Table I. It should be noted that n and U cannot be selected independently, as both determine τ_0 as well as θ. The tabulated values were also used to obtain the full lines in Fig. 8, appertaining to the temperature dependence of the activation volume in the Nb–Mo system, using Equation 2.

With the alloys considered, values of A (Equation 3) were comparable with T_2 (298 K), and it may be readily shown that this implies a near-linear relation for $\Delta\tau/\tau_1$ with a slope of

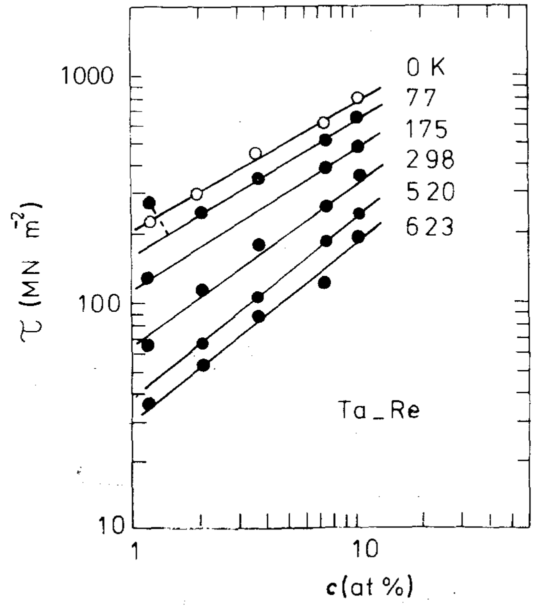

Figure 2 Concentration dependence of the CRSS, τ, of the alloys referred to in Fig. 1. The points for 0 K correspond to $\tau = \tau_0$, as given by the intersections of the curves with the ordinate in Fig. 1.

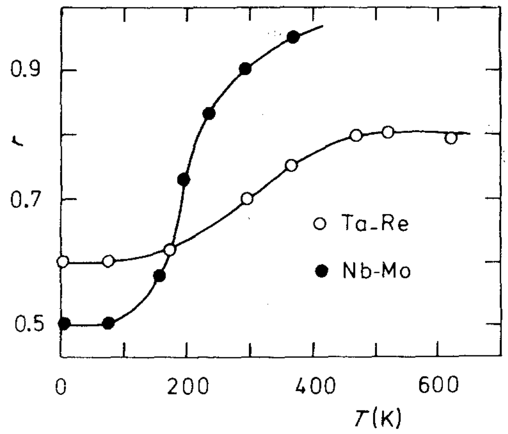

Figure 3 The temperature dependence of the exponent in the relation $\tau \propto c^r$, for the Ta–Re and Nb–Mo alloys referred to.

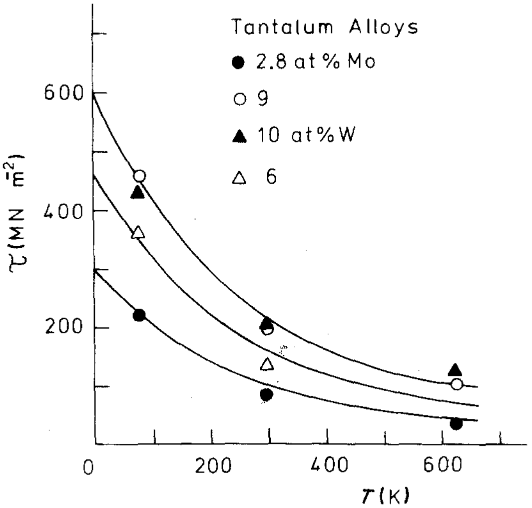

Figure 4 The CRSS of Ta–Mo and Ta–W alloys. The curves were obtained by means of Equation 1 with parameters given in Table I; the experimental points were taken from Mitchell and Raffo [6].

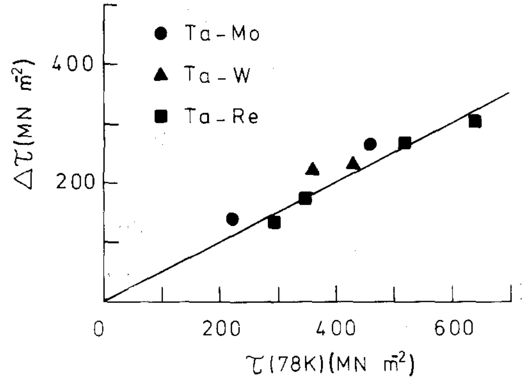

Figure 5 The dependence of $\Delta\tau \equiv \tau_1 - \tau_2$ on τ_1 for the tantalum alloys referred to in Figs. 1 to 4. $T_1 = 77\,\mathrm{K}$, $T_2 = 298\,\mathrm{K}$. As with other alloys [3], the experimental points fall close to a straight line with slope of about 0.5, as required by Equation 3.

about 0.5, as had been drawn through the data points in Fig. 5. The model suggests that for a given temperature the CRSS should be proportional to c^r with $\frac{1}{2} \leqslant r \leqslant 1$ [4, 8], the larger r values occurring at higher temperatures. Figs. 2 and 7, based respectively on the theoretical curves in Figs. 1 and 6, show that a power-law of this type is quite well applicable in both cases; equally, the values of r (Fig. 3) obtained from the slopes of the lines in Figs. 2 and 7, lie within the expected range.

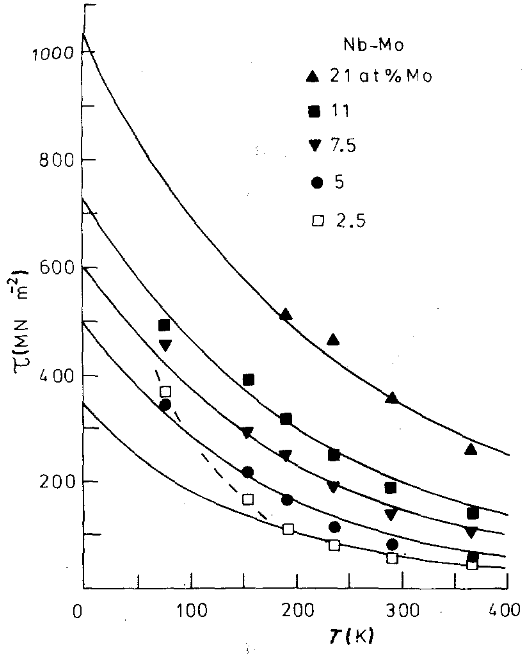

Figure 6 The CRSS of Nb–Mo single crystals. The curves were obtained as in Fig. 1; the experimental points were taken from Peters and Hendrickson [7].

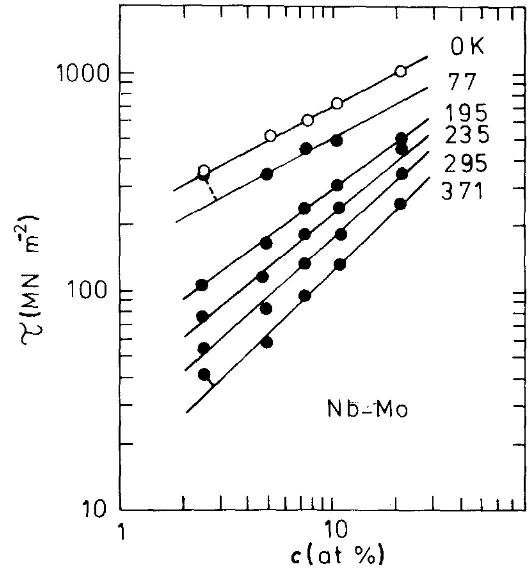

Figure 7 The τ/c-relation for the alloys referred to in Fig. 6.

3. Atypical alloys and pure metals

Data, parallel to those above, referring to the Ta–Nb and Nb–Ta alloys (Figs. 9 to 11), in which the curves through the experimental points were obtained as before, with parameters given in

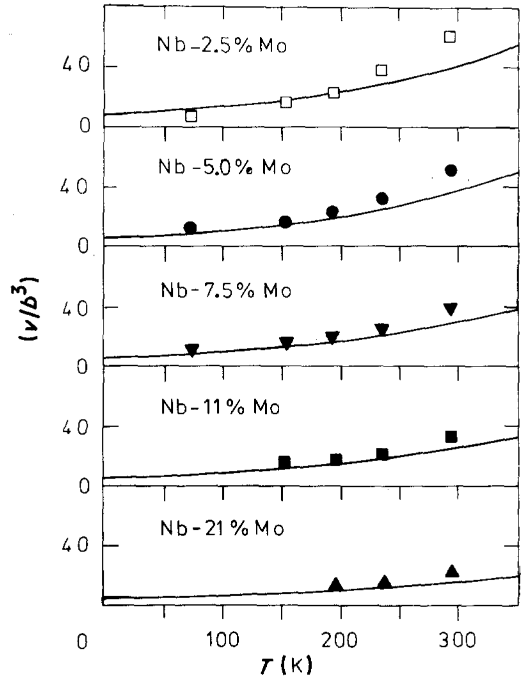

Figure 8 The temperature dependence of the activation volume for the alloys referred to in Fig. 6. The curves were obtained by means of Equation 2 with data given in Table I; experimental points were taken from Peters and Hendrickson [7].

TABLE I Parameters used in Equation 1 to obtain the curves shown in Figs. 1, 4, 6 and 8. Data for tantalum alloys from Mitchell and Raffo [6], with $G = 8.2 \times 10^4$ MN m^{-2} and $Gb^3 = 11.7$ eV; data for Nb–Mo from Peters and Hendrickson [7], with $G = 5.8 \times 10^4$ MN m^{-2} and $Gb^3 = 8.3$ eV

Alloy	c (at %)	τ_0 (MN m^{-2})	n	U (meV)
Ta–Re	1.2	220	1.5	107
	2.1	300	1.5	111
	3.8	450	1.4	115
	7.5	600	1.4	110
	10.5	800	1.3	114
Ta–W	6.0	460	1.2	80
	10.0	600	1.3	88
Ta–Mo	2.8	300	1.3	83
	9.0	600	1.3	93
Nb–Mo	2.5	350	1.0	79
	5.0	500	1.0	79
	7.5	600	1.0	78
	11.0	725	1.0	78
	21.0	1025	1.0	79

Table II show that, again, Equations 1 and 2 provide a suitable basis for the representation of the experimental results. However, a feature of particular interest arises here from the equality of the values of b (2.86A) of tantalum and niobium [11], as a result of which solid-solution hardening appears to be negligible; in fact the alloys behave as the solvent crystals (with c taken equal to 1) with respect to the temperature and concentration dependence of both the CRSS and the activation volume. By comparison with the typical alloys (e.g. Figs. 6 and 8), the temperature dependence of τ and of v (Figs. 9 to 11) is more pronounced, akin to the corresponding behaviour of sodium and potassium (Figs. 12 and 13), the data for which are also given in Table II.

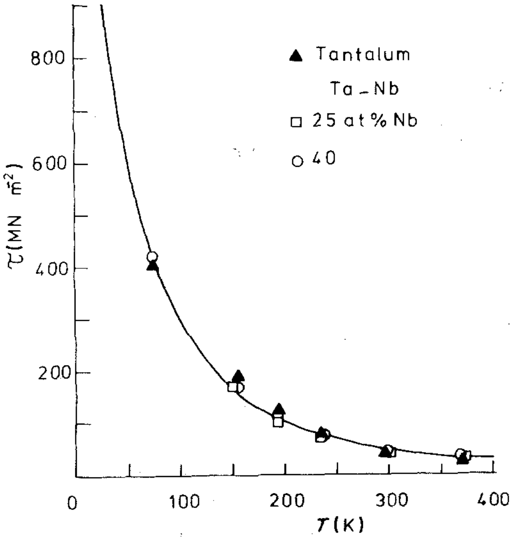

Figure 9 The CRSS of tantalum and Ta–Nb alloys. The points were taken from Michell and Raffo [6]; the line drawn through them was obtained by means of Equation 1 with $c = 1$ and the parameters given for tantalum in Table II.

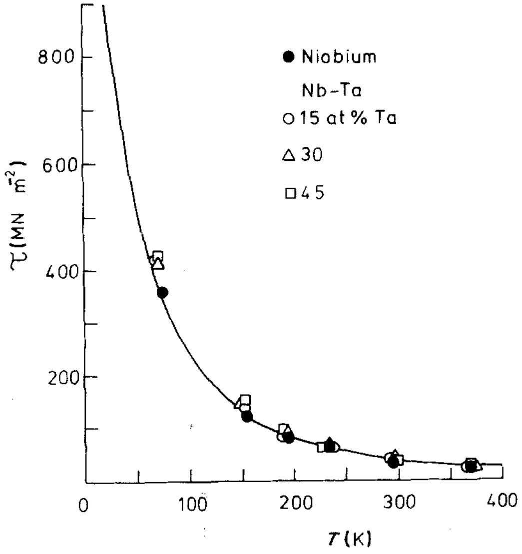

Figure 10 Nb–Ta alloys. Details as for Fig. 9.

4. Discussion and conclusions

The present results show that the relatively simple

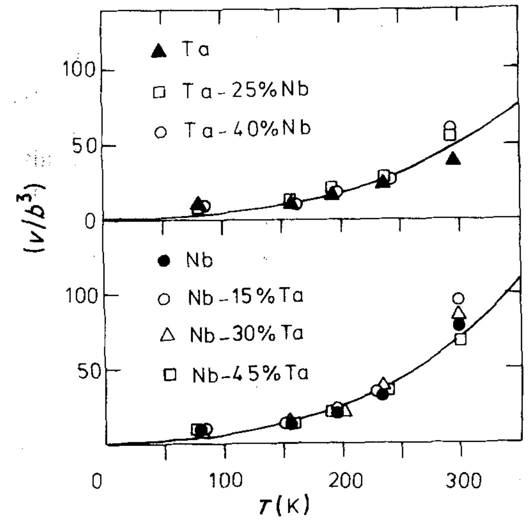

Figure 11 The temperature dependence of the activation volume for the metals and alloys referred to in Figs. 9 and 10. The curves, drawn through the experimental points taken from Peters and Hendrickson [7], comply with Equation 2, with the data given in Table II.

TABLE II Parameters used in Equation 1 to obtain the curves shown in Figs. 9 to 13, with $c = 1$ in all cases. Data for the Nb–Ta alloys were taken the same as for niobium, those for Ta–Nb as for tantalum; the pertinent references are listed

Metal	G ($\times 10^{-4}$ MN m^{-2})	Gb^3 (eV)	τ_0 (MN m^{-2})	n	U (meV)	Reference
Na	0.60	1.92	2.6	1.8	0.37	[9]
K	0.12	0.68	2.6	1.7	0.63	[10]
Nb	5.8	8.3	1400	0.3	15	[7]
Ta	8.2	11.7	1400	0.3	15	[6]

kink-pair model represented by Equations 1 to 3, and the concentration dependence of the CRSS implied by it, provide a suitable basis for the representation of the principal features of solid-solution hardening of b c c alloys. Pinning of edge dislocations by the solute atoms may render the former less mobile than the screws, so that the CRSS and the activation volume will depend on temperature and solute concentration as in f c c solid solutions. If the pinning is weak, as may result from a lack of significant interaction of the solute with dislocations, as well as from too low a solute content, "pure metal" behaviour is observed. This is also encompassed by Equations 1 to 3, with c being taken equal to 1, i.e., the equations then refer to a Peierls potential [12].

Deviations from the requirements of Equations 1 to 3 were found at the lowest concentrations used (1.2 at% in Ta–Re, shown in Figs. 1 and 2, and 2.5 at% in Nb–Mo, as is apparent from Figs. 6 and 7). In fact, as may readily be confirmed, the curves for these concentrations are rather well represented by those for pure tantalum and pure niobium respectively; both (Figs. 9 and 10) are much steeper at temperatures below about 150 K than the τ/T-relations for the concentrated solid solutions; this suggests that the Peierls mechanism rather than the solid-solution process is rate determining with the most dilute Ta–Re and Nb–Mo alloys referred to. As a result, a type of solid-solution softening emerges, in which, at temperatures below about 100 K, the yield-stress of the tantalum alloy containing 1.2 at% Re, for example, exceeds that of the solid-solution containing 2.1 at% Re (Fig. 1). The above interpretation of this effect implies that alloying does not only enhance the locking of the edge dislocations, but also reduces the lattice friction acting on the screws, the extent of the reduction increasing with the solute content.

In the model on which Equations 1 to 3 are based the details of the process by which a screw dislocation surmounts the Peierls barrier in b c c metals are not specified as, by contrast, in more specific treatments [13]. Consequently, certain features of the yield process, such as the orientation dependence of the yield-stress are not, at present, embodied in it.

The observed $\Delta\tau/\tau_1$ ratios (Fig. 5) comply with the implications of Equation 3, and the concentration dependence of the CRSS of the Ta–Re and Nb–Mo alloys agrees with the trend expected from the model.

The present data, and other work, e.g. on the deformation of niobium crystals between 4.2 and 77 K [14], do not point to the occurrence of low-temperature anomalies in the T-dependence of the CRSS, such as occur in f c c metals and alloys; however, Kirchner [15] observed a levelling-out of the temperature dependence of the yield stress of potassium crystals below about 20 K, and Takeuchi and Maeda [16] found that the mode of slip in high-purity tantalum crystals, studied by them in the range 0.7 to 40 K, was sensitive to the presence of small amounts of impurities. The origin of the deviations of the temperature dependence of the yield-stress from expected, classical, behaviour, possibly involving quantum effects [17], is not as yet fully elucidated.

It should be noted that the extrapolations of the τ/T-curves in the figures to temperatures below the experimental ranges, which were made in order to determine τ_0, are rather approximate, and are based on the assumption that the classical behaviour, as observed down to about 78 K, remains operative at lower temperatures. It may not therefore be germane to use these values of τ_0 to derive the Peierls stress of the metals considered; firstly, the effect of small amounts of impurities on τ_0 has not been fully clarified and, further, because the model is too simple to permit quantitative estimates. Finally, experimental data on the Peierls stress seem to be equivocal [18]; thus Smialek *et al.* [19] deduced a value of 360 MN m^{-2}

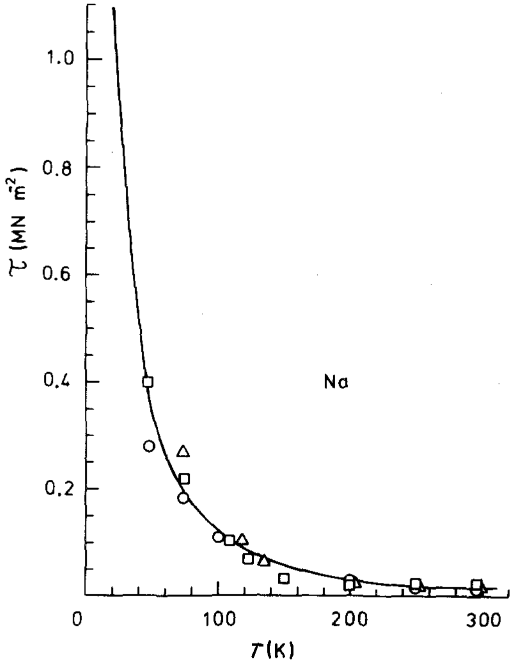

Figure 12 The CRSS of sodium single crystals. The curves, drawn through the experimental points taken from Herke *et al.* [9], comply with Equation 1, with the data given in Table II.

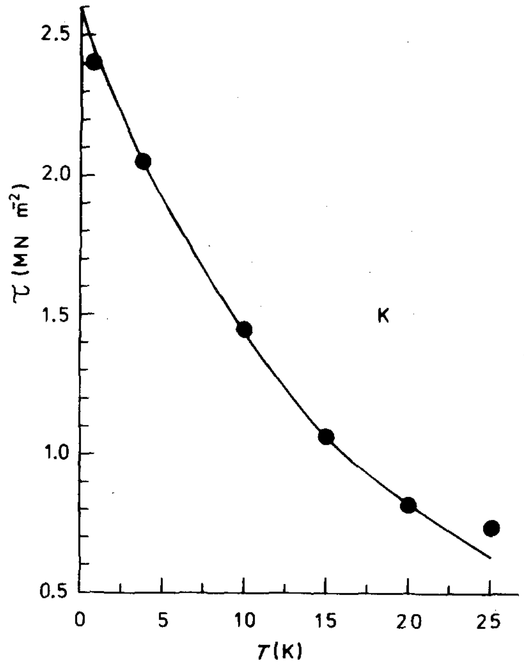

Figure 13 The CRSS of potassium single crystals. The curves, drawn through the experimental points were taken from Basinski *et al.* [10] comply with Equation 1, with the data given in Table II.

for tantalum from measurements on single crystals, i.e. substantially less than the experimentally determined yield-stress at 78 K of the tantalum crystals referred to in Fig. 9.

The model correlates the activation volume with temperature quite well (Figs. 8 and 11); some deviations at relatively high temperatures (Fig. 8) have also been observed with f c c alloys [8]. It is possible that recovery, facilitated by the movement of point defects, is responsible for the effect. The discrepancies are not removed on representing the activation volume as a function of stress rather than temperature [8], using Equations 1 and 2. Nevertheless, the agreement with the model is substantially better than is afforded by a simple power-law, e.g. $v \propto \tau^{-1/2}$ [20], obtainable from Equation 2 by omission of the second term, particularly in the case of the pure b c c metals (Fig. 11).

The similarities in the behaviour of sodium and potassium are also of interest (Figs. 12 and 13) for, as a result of phase changes, sodium would not usually be regarded as typical of b c c metals at low temperatures [21].

The values used for n and U for any of the groups of alloys listed in Table I vary but little with concentration, so that all the curves for a given type of solid-solution could quite well be described with one set of values of U and n. However, while Equations 1 to 3 can be seen to provide a good basis for representing the effect of solute content and temperature on the yield-stress of concentrated b c c solid solutions, they would not necessarily be expected to apply to the flow stress in substantially work-hardened alloys for, as in f c c solid-solutions [22] the solute may then become less effective as a source of mechanical strength than would work-hardening.

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