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INTERNAL FRICTION CAUSED BY THE INTRINSIC PROPERTIES OF DISLOCATIONS

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Abstract.— Relaxation phenomena attributable to the intrinsic properties of dislocations are reviewed. The principal theoretical models for the intrinsic dislocations relaxations and the Snoek-Köster relaxation are presented in outline. Recent important experimental results, mainly on f.c.c., h.c.p. and b.c.c. metals, are examined and compared with the theoretical predictions.

1. **Introduction.**— In 1949, Bordoni (1) observed an internal friction (IF) peak in f.c.c. metals which could only be attributed to dislocations and was associated with double-kink generation (DKG) by Seeger (2). From the relaxation enthalpy of Bordoni peaks, estimates of the Peierls stress can be obtained. The relatively high value found for the f.c.c.'s has persuaded several authors that the Bordoni peaks should not be attributed to DKG.

The objective of this paper is to review the experimental properties of the Bordoni relaxation (BR) in the f.c.c.'s together with their analogues in other structures, and to compare them with theoretical predictions. In the f.c.c.'s the situation seems clear and in the b.c.c.'s we can report the resolution of a long standing controversy about the role played by H interstitials in the IF spectra of deformed samples. However, in the h.c.p.'s and b.c.c.'s some contradictions still exist.

2. **Theoretical Aspects.**— Several models for the BR have been proposed and their essential characteristics are reviewed in detail elsewhere (3). Only two of them are generally retained to explain experimental data on the BR : DKG and geometrical kink migration (GKM). In the b.c.c.'s, the IF spectra of deformed samples of Mo, W and α -Fe are markedly different from those of samples of V, Nb, Ta and hydrogen charged α -Fe. This difference, which until recently has been the subject of a controversy, is attributable to the presence of H interstitials with some dislocation core mobility giving rise to Snoek-Köster (S-K) peaks. According to some authors (4,5), at the atomic level, the S-K(H) relaxation is due to the modification of the DKG and GKM processes by the core mobile interstitials. Thus, it is important to outline the models of the S-K relaxation as well.

2.1 **The DKG model.**— The Seeger-Paré theory (2,6) gives a good description of this model. Consider a dislocation segment (1) parallel to a Peierls valley (fig. 1). At 0K, the Peierls stress (σ_p) is necessary to displace it. At $T > 0$, under the combined influence of thermal fluctuations and the applied stress (σ_a), DKG can occur. The IF due to DKG attains its maximum when the imposed frequency (f) is of

the same order of magnitude as the frequency (ν) of DKG :

Fig. 1 : Left : schematic diagram of a dislocation overcoming a Peierls barrier by creation of a double-kink : free energy of the double-kink as a function of kink separation x .

a) initial position of the line. b) the double-kink is created. c) the kinks glide along the line.

$$\nu = \nu_l \exp(-2W_k/kT) \quad [1]$$

where $2W_k$ = formation energy of a kink pair, and ν_l is the eigenfrequency of the dislocation line.

In Seeger's model, the segment can be in either of two configurations (a) or (c). But with $\sigma = 0$, configuration (c) has an energy $2W_k$ greater than that of (a) (see fig. 1). Consequently, the occupation probability of (c) is very small and no relaxation peak will be observed. The relaxation strength only becomes observable when the exchange of dislocations between the two configurations is significant (i.e. if the two configurations are energetically equivalent).

For $\sigma \neq 0$, the two configurations can become equally probable when the Paré condition

$$|\sigma a| \approx 2W_k \quad [2]$$

is fulfilled. When σ_a is weak, this condition can only be satisfied by the internal stress (σ_i).

This theory accounts for the primary characteristics of the BR (2) but fails to explain the extent of peak broadening, the variation of peak temperature (T_M) with microstructural state and non-linear effects (strain amplitude dependence and effects of bias stress). By including kink migration, Engelke (7) has shown that some of these secondary features can be explained. However, Esnouf and Fantozzi (3,8) have proposed a new approach to the calculation of IF caused by DKG. The energy diagram of a segment lying in a Peierls valley is determined as a function of its position for different values of σ and l (fig. 2). The activation volume (v_C) and energy (E_C) for double-kink nucleation depend only on σ :

$$[3] \quad E_C = 2W_k \left[1 - \left(\frac{\sigma}{\sigma_p} \right)^n \right] \quad \begin{matrix} v_C^m \sigma = \text{const.} \\ n \approx 3/4 \quad m \approx 4 \end{matrix}$$

Fig. 2 : Potential Energy of the dislocation segment as a function of the number of double-kinks.

Kinetic theory is used to calculate the dislocation motion. The variation of the fraction (n_i) of dislocations lying in the i -th valley is given by :

$$\frac{dn_i}{dt} = \Gamma_{i-1} n_{i-1} - (\Gamma_{i,i-1} + \Gamma_{i,i+1}) n_i + \Gamma_{i+1,i} n_{i+1} \quad [4]$$

$$\sum n_i = 1$$

where the jump frequency from valley i to $i+1$ is $\Gamma_{i,i+1} = \nu_L \exp(-E_{i,i+1}/kT)$ and $E_{i,i+1}$ etc are as specified in Fig. 2. The numerical solution of the above system, for an oscillating applied stress, allows calculation of the area swept-out by the dislocation and hence the IF. All of the character-

istics of the BR are explained by this model, without taking into account kink migration. By obtaining an effective dislocation mobility for DKG, Schlipf and Schindlmayr (9) have obtained similar results.

2.1.1 Peak height.- For a population of equal segments, when the Paré condition is fulfilled, the intensity of the BR varies as ℓ^n with $n \approx 1$ for short segments and $n \approx 2$ for long segments. For distributions of ℓ and σ_i , the peak height varies as ℓ^2 for short segments. When the Paré condition is not satisfied, the variation of peak height is very pronounced. The exponent 2 shows that the area swept-out by the segment is controlled by the line tension (E_L). In addition, the relaxation strength decreases with vibration frequency.

2.1.2 Peak temperature.- T_M depends upon σ_i and ℓ (fig. 3). It is only weakly dependent on ℓ for a distribution in ℓ only, but changes significantly for distributions of both ℓ and σ_i .

Fig.3 : Variation with mean deviation in internal stress for different mean normalized lengths L_0^1 ($L_0^1 \approx 10^{-2} L/a$) :

These results show that the variation of the pre-exponential factor (τ_0) of the relaxation time with ℓ is given by : $\tau_0 \sim \ell^p$ ($0 < p < 2$). The activation energy is near $E_c(\sigma_i)$.

- a) of the relative peak height
 b) of the peak temperature T_M' (3).

2.1.3 Broadening factor.- Distribution of τ_0 and/or the activation energy cause peak broadening. It can be up to four times the factor for a Debye peak. Furthermore, the high temperature part of the peak (due to the longest segments) is more developed than the rest of the peak.

2.1.4 Non-linear effects.- The model predicts important non-linear effects. When σ_i is too small to satisfy the Paré condition, the peak can still develop and pass through a maximum as the applied stress (oscillating or static bias) increases (fig. 4). Moreover, the peak shifts to lower temperature and broadens with increasing σ_a . Similarly, the relaxation strength presents a maximum as a function of σ_i (Fig. 3). The increase in peak height is due to the Paré condition, the decrease results from the reduction of swept-out area when the number of double-kinks increases.

Fig. 4.: Variation of peak height δ'_M and temperature T'_M as a function of σ_a (for $L' = 7$ and $\sigma_i = 0$) (8).

2.2 The GKM model.- Geometrical kinks moving along close-packed directions can

be thermally activated over the second order Peierls barriers giving rise to an IF phenomenon as shown by Brailsford (10) and Müthrich (11). An applied stress causes GKM (fig. 5) given by :

Fig. 5 : Geometrical kink motion on a dislocation making an angle ψ with a close-packed direction.

$$\frac{\partial u}{\partial t} - D_k \frac{\partial^2 u}{\partial y^2} = \sigma a^2 b m n_k \quad |5|$$

$$m = D_k/kT = v_d b^2 \exp(-W_M/kT)$$

where m = kink mobility, v_d = Debye frequency, W_M = kink migration energy

and n_k the number of geometrical kinks per unit length.

Solution of the above equation yields the relaxation strength (Δ) and time (τ) of the IF peak :

$$\Delta = 4 \mu \frac{a^2 b^2}{\pi^2 kT} \Lambda \ell^2 n_k ; \quad \tau = \frac{\ell^2}{v_0 b^2 \pi^2} \exp (W_M/kT) \quad | 6 |$$

μ is the shear modulus and Λ the dislocation density. The ℓ^2 dependence of is due to behaviour analogous to a stretched string behaviour (3).

The GKM model can account for most of the SR characteristics. Distributions of ℓ and Ψ (fig. 5) lead to broadening of ~ 1.5 . This is due to distribution of τ but not the activation energy (unlike DKG). Another important prediction is that τ_0 ($\geq 10^4/\nu$) for GKM is two orders of magnitude higher than τ_0 ($\sim 10^2/\nu_0$) for DKG. Unlike DKG, a well-defined stress level is not required to observe a decrease in Δ when stress ($\sigma > \frac{\mu b \sin \Psi}{\ell}$) induces a pile-up of geometrical kinks.

2.3 The S-K relaxation.— S-K(X) relaxations where X is an impurity interstitial (II) are well established in the IF spectra of deformed b.c.c.'s. A phenomenological model proposed by Schoeck (12) has been used with varying degrees of success to explain the experimental results. It is based on the dragging of II's by stringlike dislocation segments (fig. 6a). The relaxation time (τ_{SK}) and strength (Δ_{SK}) found using this model for a dilute concentration (C_d) of II's on the dislocations are given by :

$$\tau_{SK} = \alpha kT L_0^2 C_d \exp (H_{SK}/kT) ; \quad \Delta_{SK} = \beta \Lambda L_0^2 \quad | 7 |$$

and the relaxation enthalpy (H_{SK}) by : $H_{SK} = H^B + H^{M'} \quad | 8 |$

In these equations, L_0 is the mean segment length, H^B the binding enthalpy of the II's to the dislocations, $H^{M'}$ the migration enthalpy of the II's at the dislocations (H^M is the same quantity in the lattice) and α, β are numerical constants. When the cores are saturated, H_{SK} reduces to $H^{M'}$. Recently,

Fig. 6 : Models of the S-K relaxation : a) string model. Verдини and Bacci (13) have developed a modification of the model which

takes into account the dissipation associated with bow-out between the II's. They show that Δ_{SK} becomes dependent on C_d (below saturation) and that $H_{SK} = 2H^B + H^{M'}$. Other modifications have been reported. For example, Miner et al. (14) have shown that when the occupation probabilities of the pinning sites are more correctly treated by Fermi-Dirac statistics, H_{SK} can take on values between $H^B + H^{M'}$ and $2H^B + H^{M'}$. From many comparisons of the predictions of Schoeck's model with experiment, there is little doubt that several aspects of the model are correct.

Seeger (4) attributes S-K (X) to DKG in the presence of core mobile X interstitials (fig. 6b). In this model, with the exception of the numerical constants, eqn. |7| is unchanged, while the relaxation enthalpy becomes :

$$H_{SK} = 2 H_k + H^{M'} \quad |9|$$

where $2 H_k$ is the formation enthalpy of a kink pair. Hirth (5) considering the special case of S-K(H) in α -Fe, has developed a similar model for the case of near saturation. He finds

$$H_{SK} = 2 H_k + H^{M'} - (\mu \sigma b^6 / 2\pi)^{1/2} \quad |10|$$

At the stress levels normally involved eqns. |9| and |10| become practically identical. However, the last term does give an explicit dependence on σ (and hence σ_1) which can account for the broadening of the S-K peaks. The most important point to establish is which of the predictions, eqn. |8| (or its modifications) or eqn. |9| (or |10|) comes closest to representing the experimental values of H_{SK} .

3. Important Experimental Results.— Our task in this section is to examine the important experimental results on the intrinsic relaxations of dislocations in metals and to try to interpret them with the models outlined in section 2. Results published up to the time of the Manchester conference (1979) have been reviewed recently (3), so we will concentrate mainly on more recent results.

3.1 F.C.C. and H.C.P. metals.— The BR in these metals is usually composed of two IF peaks : B_1 (the Wiblett and Wilks peak) and B_2 (the Bordoni peak). In metals of ordinary purity the BR is not observed in annealed specimens and only appears after plastic deformation. In ultra high purity (UHP) samples, the BR is always observed, even after recrystallization, but depends strongly on the applied stress. This is shown on fig. 7 for Pb (15) and on fig. 8 for Ag (16). Thus the disappearance of the BR in non-UHP metals after recrystallization is not due solely to the decrease in Δ but also to pinning by impurities. This shows that the BR can only be attributed to the intrinsic properties of dislocations, i.e. GKM and/or DKG.

Fig. 8 : Evolution of the BR during recrystallisation at 200°C in 6N silver cold-rolled 6% at 20°C ($\epsilon_m = 2.5 \times 10^{-7}$) : 1, 0mn; 2, 10mn; 3, 30mn; 4, 60 mn; 5, 120mn; 6, 840mn (16).

Fig. 7 : Evolution of the BR in Pb : 1, unannealed; 2, annealed at 470K; 3, at 550K (15).

3.1.1 Peak height.— The peak height varies with impurity content, conditions of plastic deformation, annealing and irradiation. This variation is linked to the evolution of ℓ and is of the form ℓ^r with $1 \lesssim r \lesssim 2$.

3.1.2 Peak temperature.— The peak temperature varies slightly with the microstructural state and is caused essentially by a dependence of τ on ℓ ($\propto \ell^s$ with s lying between 1 and 4) (3,17). This variation explains the large dispersion of the data in Arrhenius plots.

3.1.3 Broadening factor.— The peaks are generally two to four times broader than a Debye peak, i.e. τ is distributed. They are broader for small amounts of plastic deformation, with the increase in width more pronounced on the high temperatures sides. In contrast, annealing decreases the widths by reducing the high temperature sides preferentially. This asymmetry in the reduction of the width is shown by the differences in the IF curves for slightly deformed Al in fig. 9 (18).

Fig. 9 : the difference $\delta_n - \delta_{n+1}$ between the curve obtained during the linear warm-up, n , and that obtained during the linear warm-up, $n+1$, after flexural deformation of 0.2 % at 80 K ($n = 3$, annealing temperature $T_R = 222$ K; $n = 4$, $T_R = 248$ K; $n = 5$, $T_R = 285$ K; $n = 6$, $T_R = 314$ K). The short arrow indicates the peak temperature of B_2 , while the long arrow indicates the temperature of the maximum of $\delta_n - \delta_{n+1}$ (18).

3.1.4 Non-Linear Effects.- When σ_i is significant (after large plastic deformation), σ_a has little effect (fig. 10). However, for weak σ_i

Fig. 10.: Internal friction in gold cold-worked 7 % at 20°C : a) after deformation, b) after annealing 60 min at 350°C. Oscillating strain amplitude : (1) $\epsilon_m = 4 \times 10^{-7}$, (2) 15×10^{-6} , (3) 6×10^{-6} , (4) 5×10^{-5} . c) Internal friction in gold cold-worked 62 % and annealed 60 min at 165°C. (1) $\epsilon_m = 3 \times 10^{-7}$, (2) 1.5×10^{-6} , (3) 6×10^{-6} , (4) 1.5×10^{-5} (19).

important non-linear effects are observed (Fig. 10): the relaxation strength depends strongly on σ_a (periodic or static). Similar results have been observed by many authors (3). The BR intensity increases considerably with the vibration amplitude or static stress. Moreover, the removal of the BR during recovery or recrystallization is caused by a decrease of σ_i . In some cases, the peak height passes through a maximum as a function of the strain amplitude. In Mg, significant non-linear effects are observed after plastic deformation when only one slip system is activated (20). As shown in fig. 11, the peak heights of both B_1 and B_2 pass through a maximum as the superimposed bias stress is increased.

Fig. 11 : Variations of the peak heights of B_1 and B_2 as a function of bias stress σ_s ($\sigma_m = 5 \times 10^{-6}$); 1, $\sigma_s = 0$; 2, $\sigma_s = 8 \times 10^{-6}$ μ ; 3, $\sigma_s = 2 \times 10^{-5}$ μ ; 4, $\sigma_s = 5 \times 10^{-5}$ μ ; 5, $\sigma_s = 10^{-4}$ μ .

B_1' and B_2' : after subtraction of the background (20).

3.1.5 Low Temperature Modulus Defect.- In both f.c.c.'s and h.c.p.'s, there is a modulus defect at temperatures lower than the BR. Consequently, GKM, on at least one of the dislocation types, must be reserved to explain this phenomenon.

3.1.6 Dislocation types.- Seeger (2) proposed that non-screw dislocation segments ($\perp$'s) give rise to B_1 and screws ($\parallel$'s) give rise to B_2 , since the Peierls energy for $\parallel$'s is higher than that for $\perp$'s. This is supported by observations that B_1 is more sensitive to pinning by point defects than B_2 . However, recent electron microscopy studies by Lauzier and Minier (21) support an earlier hypothesis of Thompson and Holmes (22), i.e. B_1 is associated with 90° and 30° dislocations lying along $\langle 112 \rangle$ direction and B_2 is associated with 60° and screw dislocations lying along $\langle 110 \rangle$ direction. The $\parallel$'s are responsible for the high temperature component of B_2 whereas the 60° dislocations are responsible for the low temperature component. This hypothesis allows us to explain the variation of

the peak temperature during irradiation or annealing (21) and also accounts for the two peaks X and Y observed in deformed Ni (23) : peak X may be the analogue of B_2 and Y the analogue of B_1 . Besson (23) proposed that Y was caused by dissociated forms of the same dislocations responsible for X. The Thompson and Holmes hypothesis is more plausible and does not necessitate the assumption of an increase in dissociation with plastic deformation. Nevertheless, the rôle of stacking fault energy in the BR remains to be elucidated. In Mg, Seyed-Reihani (20) has shown that results lead to a conclusion in keeping with Thompson and Holmes hypothesis also.

3.1.7 Comment.- Although the BR has been observed in several h.c.p.'s, only in UHP Mg (20) have B_1 and B_2 peaks with properties comparable to those in the f.c.c.'s been reported. We conclude that f.c.c.'s and h.c.p.'s behave similarly (3). Nevertheless, further studies on other UHP h.c.p.'s will be necessary to confirm this point.

3.1.8 Interpretation.- Only the DKG model accounts for all of the experimental characteristics of the BR as can be seen from a direct comparison of sections 2.1.1. to 2.1.4. with sections 3.1.1. to 3.1.4. Nevertheless, the value of σ_p obtained from the relaxation enthalpy of the BR (10^{-3} - 10^{-4} μ) is considerably greater than that predicted from plastic behaviour and so, the DKG model for the BR is still contested (3,17). Two relatively new types of experiment also support the DKG model. Firstly, measurements of ultrasonic attenuation at very low temperatures demonstrate GKM (24) : the attenuation decreases as the bias stress increases (fig. 12). This "stiffening" phenomenon is due to the pile-up of geometrical kinks

Fig. 12 : Variation of attenuation versus stress at different testing temperatures in Al (24).

in response to the bias stress (25). In the temperature range of the BR, dislocation segments surmount the Peierls barriers and the attenuation increases with bias stress (fig. 12). Some different results were reported recently (26) but remain to

be confirmed. Secondly, a thermally activated microdeformation stage (3,27) associated with the BR has been observed in Al, Au and Mg (fig. 13) together with a variation in the anelastic limit (27). A quantitative analysis of these results yields an activation energy of ~ 0.1 eV and a volume of $\sim 100 b^3$ entirely in keeping with the DKG model (27).

In conclusion, the experimental results strongly support the existence of a Peierls-Nabarro stress in the f.c.c and h.c.p metal of 10^{-3} to 10^{-4} μ .

Fig. 13 : Microdeformation (and derivative curve) as a function of temperature for a 6N gold sample, deformed 0.7 % in torsion at 10 K and annealed at 150 K : (a) with a static stress $\sigma = 5 \times 10^{-4}$, (b) after suppression of the stress at 10 K (3).

3.2 B.C.C. Metals.— The intrinsic dislocations relaxations in the b.c.c.'s are of special interest because the mobilities of $\perp$'s and Θ 's are such that the analogue of the Bordoni relaxation is split into two well-separated peaks (28,29) : a low temperature peak associated with DKG ($\perp$) and a peak at a much higher temperature associated with DKG (Θ). Moreover, geometrical kinks on Θ 's are expected to be rather abrupt (effectively high second order Peierls barriers (30)) so that GKM (Θ) should fall into a temperature range close to that for DKG ($\perp$). As in the case of the f.c.c.'s, GKM ($\perp$) is expected to occur at very low temperatures (< 4 K) and manifest itself as a modulus defect at the lowest temperatures investigated. These expectations seem to be realized in Mo, W and α -Fe as typified by the IFS of a deformed single crystal of Mo (31,32) in fig. 14. In contrast, in V, Nb

Fig. 14 : Internal friction spectrum of a deformed crystal of Mo. Plastically strained 1 % at 200 K + a small additional plastic strain in torsion at 295 K (31).

and Ta (also α -Fe + H) the IFS is characterized by a larger number of peaks. Typical re-results for a deformed single crystal of Ta (33) are shown in fig. 15 where the IFS is resolved into 7 component peaks with only one of them attributed (by the authors) to the intrinsic dislocation processes discussed above; the others are attributed to the interactions of dislocations with Π 's.

Fig. 15: Internal friction spectrum of a deformed single crystal of Ta. Deformation history : 3.1 % tensile strain at 375K + 0.5 % tensile strain at 200K + a small additional plastic strain in torsion at 295K (33).

Most of the differences between the f.c.c.'s and b.c.c.'s can be traced to the strong interactions between residual II's (H, N, O or C atoms) with the dislocations in b.c.c.'s. In fact, these interactions can lead to diametrically opposed changes in properties, e.g.

hardening through the interactions of II's with $\perp$'s and softening through their interactions

with Θ 's (see review by Pink and Arsenault (34)). The presence of trace amounts of H in deformed V, Nb and Ta samples has led to a controversy over the interpretation of the IFS in these metals which has been resolved recently.

3.2.1 A Controversy Resolved.— The IFS of deformed V, Nb or Ta is generally characterized by a group of relaxations in the range 120-200K (e.g. fig. 15) labelled α by most workers (35,36). However, a peak labelled δ , at even lower temperatures, is observed in some samples (35,37). As suggested by Bruner (38), the Rome school (35,37) has argued that α requires the presence of both H and dislocations and that δ can be attributed to intrinsic dislocation processes. This interpretation was contested by the Stuttgart School who studied samples with considerably lower II concentrations and observed neither δ nor any indication of dislocation movement below α , leading them to attribute α to intrinsic dislocation processes. Nevertheless, over the last few years, evidence that α involved H began to accumulate : notably from Shibata et al. (39) and Mizubayashi et al. (40) for V, Ferron et al. (41), Kuramochi et al. (42) and Verdini and Bacci (43) for Nb and Rodrian (33) for Ta. Yet, no explanation for the absence of δ in samples with lower II (particularly O) contents was offered. More recently, work on UHP Nb (31, 44, 45) has demonstrated that α involves H and that the major component can be attributed to an S-K(H) relaxation (43). In the highest purity samples ($C_H \ll 1$ at. p.p.m., $C_O < 5$ at. p.p.m), the work at Stuttgart (31, 44, 45) has shown that another peak appears below α (~ 70 K as shown in fig. 16). But, in order to completely settle the controversy, it is necessary to show that the 70 K peak is equivalent to δ . In an attempt to do this, Maul (45) repeated measurement on sam-

ples doped with O and H ($C_O = 800 - 1000$ at, p.p.m and $C_H = 20 - 30$ at. p.p.m). The results are shown in Fig. 17 and compared directly with δ (35).

Fig. 16 : Internal friction of a single crystal of Nb after a thorough degassing ($C_H \ll 1$ at. p.p.m) and deformation at 20 K (45).

Fig. 17 : Internal friction in a doped Nb single crystal ($C_O = 800 - 1000$ at. p.p.m). 1, after deformation 2 after H charging to $C_H \approx 20-30$ at. p.p.m; 3, results of Mazzolai and Nuovo (35,45).

It is clear that in doped Nb there is a peak ~ 30 K which is equivalent to δ . A suggested explanation of these results lies in the trapping of mobile H at immobile O (or N) so that the number of H atoms available to decorate the dislocations is reduced, allowing the intrinsic mechanisms to be revealed in less pure samples (44, 45). However, we do not regard the equivalence of the 30 K (δ) and 70 K peaks as having been demonstrated conclusively. Since δ is observed only in the presence of significant quantities of O (or N), we suggest that δ is predominantly GKM ($\odot$) and that DKG ($\perp$), which is predominant in the 70 K peak, is pinned-out. This is supported by four types of observation : i) a shift as large as 70 K to 30 K is not observed in controlled pinning experiments, where the shift for an equivalent reduction in peak height is only 5-6 K (35,3);

ii) the relaxation parameters of δ are more consistent with GKM ($\odot$) than DKG ($\perp$) (section 3.2.4); iii) results on thoroughly outgassed V (40) appear to exhibit both

peaks in the same spectrum, while the higher temperature peak is absent in samples outgassed to a lesser extent; iv) α (attributed to DKG ($\perp$) in α -Fe) cannot be generated in the presence of only ~ 20 at. p.p.m of C or N (46).

3.2.2 S-K (H) Peaks.- It is interesting to examine the predictions of the models outlined in section 2-3 for the case of the S-K (H) in α -Fe : i) because it is the most thoroughly investigated S-K (H) and ii) because it is similar in most respects to the S-K (H) (labelled α) in V, Nb and Ta. To check the validity of eqns [8] and [9] we must compare the experimental values of H_{SK} (ranging from 0.11 to 0.33 eV) (47-50) with $H^B(\perp) + H^{M'}$ and $2 H_k(\perp) + H^{M'}$. From studies of peak α in Fe, $2 H_k(\perp)$ is in the range 0.04 to 0.06 eV (48, 50, 51), while the most probable value of $H^B(\perp)$ is 0.61 eV (52). Whether we take $H^{M'} \approx H^M$ (following Seeger (4)) or $H^{M'} \approx H^M/2$ (following Hirth (5)), since the values of H^M currently advocated by various reviewers lie in the range 0.042 to 0.069 eV (53,54), no combination of the data seems to be consistent with either the Schoeck model, eqn. [8], or the Seeger or Hirth models, eqn. [9], unless the lower bound of H_{SK} (found in a series of magnetic after-effect studies only) is compared with the upper bound of $2 H_k(\perp) + H^{M'}$ as pointed out by Kê et al. (49). We suggest that this very unsatisfactory situation arises because $2 H_k(\perp)$ is effectively increased in the presence of H. This is consistent with the viewpoint expressed by Hirth (5) who attributes the S-K (H) in Fe to a combination of hardened $\perp$'s and softened Θ 's.

3.2.3 Classification of the IF Peaks.- From the above, it is clear that the labelling of the peaks in the IFS of a deformed b.c.c. has become confusing. We propose that the labelling should be modified in the manner shown in table 1 to retain as much of the traditional classification scheme (3) as possible, yet to reflect the mechanisms involved, rather than simply the temperature ranges in which the peaks occur.

3.2.4 Other Recent Experimental Results.- From fig. 14, it is clear that in UHP deformed b.c.c.'s the IFS consists of a low temperature group of peaks and a high temperature peak. The low temperature group is generally attributed to one or both of the $\alpha'(\Theta)$ and α processes; the high temperature peak is attributed to the γ process. In the following, we refer to them simply as α'/α and γ respectively. Evidence for the presence of at least two discrete relaxations in α'/α consists of the marked asymmetry or shoulder which appears on the low temperature side of the peak and a corresponding step in the modulus curve. Four such steps have been reported for the lowest temperature peak in V (40). This is similar to the stepped modulus curves associated with the BR (3).

a) Relaxation Parameters.- Activation parameters from the peak shift of α'/α with frequency fall into two groups : for V (37), Nb (43,55) and Ta (35) the activation energy falls in the range 0.012 to 0.047 eV and the pre-exponential factor falls into the range 10^{-9} to 2×10^{-8} s; for Mo (3,32), W (3) and α -Fe (51,46), the

equivalent ranges are 0.04 to 0.29 eV and 1.4×10^{-14} to 5×10^{-10} s. The first set, i.e. the old δ peaks, seems to be consistent with GKM (θ); while the second set seems to be more consistent with DKG ($\perp$). It will be interesting to see how the relaxation parameters of the recently discovered 70 K peak in Nb and the 57 in V compare with those of the old δ peaks.

TABLE 1
CLASSIFICATION SCHEME FOR THE IF PEAKS IN
DEFORMED BCC METALS

PROCESS	OLD LABEL(S)	NEW LABEL
INTRINSIC DISLOCATION LOW TEMP. } GKM ($\perp$) } GKM (θ) } DKG ($\perp$)	α' $\alpha' \alpha_2$ α (Mo, W, Fe) δ (V, Nb, Ta) 57 K (V) 70 K (Nb)	$\alpha' (\perp)$ $\alpha'_i (\theta)$ α_i i = 1,2,3 } α'/α
INTRINSIC DISLOCATION HIGH TEMP. } DKG (θ)	$\gamma, \gamma_1 \gamma_2$	$\gamma (i)$
INTRINSIC INTERACTION	$\beta_i, i = 1, 2, 3$	$\beta_i (\perp)$ $\beta_i (\theta)$
EXTRINSIC INTERACTION CORE-MOBILITY } GKM } DKG	α (V, Nb, Ta) C.W.P. S-K	S-K' (X), S-K' (X, $\perp$) S-K (X), S-K (X, θ) (X = H, O, N or C)
EXTRINSIC INTERACTION IMMOBILE PINS	β_i P_i	$\beta_\alpha(X), \beta_\gamma(X)$

Relaxation parameters so far obtained for the γ peaks in Nb (56), Ta (33), Mo(32), W(3), α -Fe (46,57,58) are all entirely consistent with DKG (θ). The γ peak narrowness and instability is taken into account by a model proposed by Astié (58). In this model, the θ 's move by thermal activation in the internal stress field and the dislocation segments change from an initial straight position towards a bowed-out position. During this motion, the screw part of the dislocation decreases and the γ peak disappears. This model gives a good description of the γ behaviour in b.c.c.'s.

b) Conditions of Plastic deformation.— Studies of the amount and temperature of plastic deformation on the α'/α and γ peaks, particularly in Mo (32,59), W (77) and α -Fe (58, 60) indicate that a major component of α'/α is due to $\perp$'s and that γ is due to Θ 's. In α -Fe the entire α'/α complex is attributed to DKG ($\perp$) (60), which suggests that the apparent structure of α'/α is due to the presence of several types of $\perp$'s with nearly equal mobilities.

c) Effects of Impurities.— In α -Fe, the interaction of II's (other than H) with $\perp$'s gives rise to β_α (60,61). Similar peaks are expected in the other b.c.c.'s but in V, Nb and Ta they are probably inhibited by the interactions of O (or N) with residual H, or masked by the S-K (H) relaxation. Extra interaction peaks are observed in the results of Rieu (59) on single crystals of Mo and W. Of these, the peak labelled β_2 by the author is attributed to the interaction of dislocations with unspecified impurities. In Nb (62) and Ta (fig. 15), the two components γ_1 and γ_2 that are candidates for DKG (Θ) behave differently in the presence of II's. It has been reported that in Nb γ_1 requires the presence of O (62), whereas in Ta, γ_2 requires the presence of O (33). In Fe, β_γ (60, 61) seems to require the presence of C (or N). It is probable, that in all the b.c.c.'s of sufficiently high purity γ is a symmetric single peak (see fig. 14 for Mo). In substitutional Fe alloys, Astié (58) has observed two peaks called γ_I and γ_{II} (fig. 18) : γ_I appears around 230 K and γ_{II} at about 320 K. For dilute alloys, γ_I and γ_{II} appear together whereas for concentrated alloys only γ_I appears. The behaviour of these two peaks is quite similar to the γ peak in pure iron.

Fig. 18 : Internal friction spectra of Fe-Ti alloys predeformed 5 % at 300 K then deformed at 120 K : influence of Ti concentration (58).

The γ_I peak is associated to DKG on softened Θ 's and the activation enthalpy measured from IF experiments is in good agreement with the values deduced from plasticity behaviour (58). The γ_{II} peak is due to DKG on normal Θ 's.

d) Effects of Irradiation.- Low temperature, low dose irradiations of predeformed b.c.c's produce similar changes to the IFS as low temperature deformation. Both create intrinsic point defects which interact with the dislocations and give rise to interaction peaks analogous to the Hasiguti peaks in the f.c.c's (3). The most detailed study of a deformed and irradiated b.c.c. is the study of α -Fe reported by the Grenoble group (46, 60, 61). The evolution of the IFS after 20K

Fig. 19 : Evolution of the internal friction spectrum of deformed and electron irradiated (20 K) α -Fe during successive anneals (60,61).

electron irradiation is shown in fig. 19 and the results of experiments comparing the low temperature irradiation with deformation are summarized in fig. 20. Generation of intrinsic point defects at low temperatures suppresses α'/α and generates β_1 . After annealing in stage III of the recovery of electrical resistivity, the dislocations which give rise to α'/α are cleaned of their intrinsic point defects and α'/α reappears. By following changes in the relaxation intensity and peak temperature of α'/α and β_1 during annealing, stages similar to those observed in the recovery of resistivity can be observed, allowing the IFS to be used as a sensitive tool in the study of radiation damage.

In Nb and Ta, it is apparent that the interaction of traces of H with vacancies leads to further complications (33,45).

Effects of irradiation on γ have not been studied in any detail. A second interaction peak, β_γ , is observed at temperatures just below γ in α -Fe (46). It is attributed to the dragging of a self-interstitial-type defect by geometrical kinks on Θ 's.

e) Low Temperature Modulus Defect.- Changes in the dynamic elastic moduli at very low temperatures have been reported in V (< 6K) (40), Nb (< 15K) (45), Ta (< 20K) (33) and α -Fe (< 4K) (60). These results are generally attributed to $\alpha'(\perp)$ (i.e. $GKM(\perp)$). However, in Fe there is some evidence that $GKM(\Theta)$ may be involved as well (60).

f) Effects of Oscillation Amplitude and Bias Stresses.- In the Nb single crystals (42), α'/α increases in intensity with increasing strain amplitude.

Fig. 20 : Evolution of the $\alpha'/\alpha + \beta_1$ complex in deformed α -Fe following electron irradiation or supplementary cold-work at 77 K (60,61) or after neutron irradiation at 77 K (46).

Fig. 21 : (a) IF peaks α'/α and β_{α} in Fe after 2 % deformation at RT. (b) Microdeformation cycle : AB, loading to $\sigma_s = 6 \times 10^{-5}$; BC heating under stress; CD, cooling under stress; DE, unloading; EF, recovery without stress; (c) Derivative curves of the α'/α stage : 1 $\sigma_s = 3 \times 10^{-5} \mu$; 2 $\sigma_s = 6 \times 10^{-5} \mu$ (64).

In α -Fe, strain amplitude dependence occurs at 5×10^{-6} even at 4K and increases with increasing temperature (61,63). Recent studies of the strain amplitude dependence associated with α'/α and γ in Mo single crystals (32) yield activation volumes of 30-150 b^3 for the major component of α'/α and $\sim 50 b^3$ for γ . These results are entirely consistent with DKG (1) and DKG (0) respectively.

Increasing strain amplitude causes the intensity of γ in Nb to increase (62) and bias stresses increase the high temperature background. These results are analogous to those reported for the Bordoni relaxation (3).

g) Microdeformation.-

Microdeformation stages associated with α'/α (see fig. 21) and γ (58,65) have been reported recently. The results obtained using a torsion pendulum technique can be summarized as follows i) the activation volume associated with α is in the range 10 to 100 b^3

in keeping with DKG ($\perp$); ii) a substructure corresponding to α' is observed but has not been analysed in detail; iii) the recovery stage (α_r) (see fig. 20) (b)) in the absence of stress is always at a lower temperature than stage α ; iv) a micro-deformation stage is also associated with β_α but its characteristics are such that it is not amenable to the same type of analysis used in the case of stage α ; and v) the stage associated with γ (58,65) can be correlated with the zero-point-drift of the pendulum in conventional studies of the IF peak γ and the results confirm that Θ 's are responsible for both the microdeformation stage and peak γ . Furthermore, the model proposed by Asti $\acute{e}$ et al. (65) (see 3.2.4. a) can also describe the γ microdeformation in b.c.c.'s.

3.2.5 Summary.- The IF peaks α'/α and γ (at ≈ 1 Hz) are illustrated schematically (together with the S-K (H) peaks)

in fig. 22. In Mo, W and α -Fe, α'/α contains a major component due to DKG ($\perp$), whereas in V, Nb and Ta containing significant amounts of both H and O (or N) it seems that DKG ($\perp$) is pinned-out and α'/α (previously δ) is predominantly due to GKM (Θ). However, in UHP Nb a major component of α'/α is attributed to DKG ($\perp$). In V, Nb, Ta and α -Fe there is a good evidence that $\alpha'(\perp)$ occurs below the lowest temperatures yet investigated.

Fig. 22 : Schematic diagram showing the relative positions and shaped of the α'/α and γ peaks at about 1 Hz in ultra high purity b.c.c. transition metal.

Although no one b.c.c. has been studied to the extent that all the primary and secondary characteristics of the BR (3) have been confirmed for both α'/α and γ ; each of the characteristics is exhibited by one or more of the b.c.c. metals taken as a group. Now, there can be little doubt that α'/α and γ taken together contain the analogue of the BR in the f.c.c.'s.

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