INTERNAL FRICTION CAUSED BY THE INTRINSIC PROPERTIES OF DISLOCATIONS
Gilbert Fantozzi, I. Ritchie

To cite this version:
Gilbert Fantozzi, I. Ritchie. INTERNAL FRICTION CAUSED BY THE INTRINSIC PROPERTIES OF DISLOCATIONS. Journal de Physique Colloques, 1981, 42 (C5), pp.C5-3-C5-23. <10.1051/jphyscol:1981501>. <jpa-00220957>

HAL Id: jpa-00220957
https://hal.archives-ouvertes.fr/jpa-00220957
Submitted on 1 Jan 1981

HAL is a multi-disciplinary open access archive for the deposit and dissemination of scientific research documents, whether they are published or not. The documents may come from teaching and research institutions in France or abroad, or from public or private research centers. L’archive ouverte pluridisciplinaire HAL, est destinée au dépôt et à la diffusion de documents scientifiques de niveau recherche, publiés ou non, émanant des établissements d’enseignement et de recherche français ou étrangers, des laboratoires publics ou privés.
INTERNAL FRICTION CAUSED BY THE INTRINSIC PROPERTIES OF DISLOCATIONS

G. Fantozzi and I.G. Ritchie*

INSA, 69621 Villeurbanne, France
*A.E.C.L.-W.N.R.E. Pinawa, Canada

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 OK, the Peierls stress ($\sigma_p$) is necessary to displace it. At $T > 0$, under the combined influence of thermal fluctuations and the applied stress ($\sigma_a$), 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 \( \nu \) of DKG:

\[
\nu = \nu_k \exp \left(-2 \frac{2W_k}{kT}\right) \quad |1|
\]

where \( 2W_k \) = formation energy of a kink pair, and \( \nu_k \) 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

\[
\sigma \bar{a} b = 2W_k \quad |2|
\]

is fulfilled. When \( \sigma_a \) is weak, this condition can only be satisfied by the internal stress \( \sigma_0 \).

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 \( T_H \) 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 (3,6) 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 \( \sigma \) and \( \beta \) (fig. 2). The activation volume \( (v^a) \) and energy \( (E^a) \) for double-kink nucleation depend only on \( \sigma :\)

\[
|3| \quad E^a = 2W_k \left[1 - \left(\frac{\sigma}{\sigma_p}\right)^n\right] \quad v^a \sigma = \text{const.} \quad n = 3/4 \quad m = 4
\]
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:

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

\( \sum n_i = 1 \)

where the jump frequency from valley \( i \) to \( i+1 \) is \( \Gamma_{i,i+1} = v_0 \exp(-E_{i,i+1}/kT) \) 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 characteristics of the BR are explained by this model, without taking into account kink migration. By obtaining an effective dislocation mobility for DKS, 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 \( \xi^n \) with \( n = 1 \) for short segments and \( n = 2 \) for long segments. For distributions of \( \xi \) and \( \sigma_i \), the peak height varies as \( \xi^2 \) 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 \( \sigma_i \) and \( \xi \) (Fig. 3). It is only weakly dependent on \( \xi \) for a distribution in \( \xi \) only, but changes significantly for distributions of both \( \xi \) and \( \sigma_i \). These results show that the variation of the pre-exponential factor \( (\tau_0) \) of the relaxation time with \( \xi \) is given by: \( \tau_0 \sim \xi^p \) \( (0 < p < 2) \). The activation energy is near \( E_c (\sigma_i) \).

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

Fig. 3: Variation with mean deviation in internal stress for different mean normalized lengths \( L_0 \) \( (L_0 = 10^{-2} L/a) \):
a) of the relative peak height
b) of the peak temperature $T_i^i$ (8).

2.1.3 Broadening factor.- Distribution of $\tau_0$ 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 $\sigma_i$ 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 $\sigma_a$. Similarly, the relaxation strength presents a maximum as a function of $\sigma_i$ (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 $\delta_M^i$ and temperature $T_i^i$ as a function of $\sigma_a$ (for $L^i = 7$ 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 Sraiford (10) and Müthrich (11). An applied stress causes GKM (fig. 5) given by:

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

$$m = \frac{D_k}{kT} = \nu_d b^2 \exp \left( - \frac{W_M}{kT} \right)$$

where $m = \text{kink mobility}$, $\nu_d = \text{Debye frequency}$, $W_M = \text{kink migration energy}$
and \( n_k \) the number of geometrical kinks per unit length.

Solution of the above equation yields the relaxation strength \( (\Delta) \) and time \( (\tau) \) of the IF peak:

\[
\Delta = 4 \mu \frac{a^2 b^2}{\pi^2 kT} \Lambda^2 n_k; \quad \tau = \frac{x^2}{\nu D b^2 \pi^2} \exp \left( \frac{W_M}{kT} \right)
\]

where \( \mu \) is the shear modulus and \( \Lambda \) the dislocation density. The \( x^2 \) dependence of \( \tau \) is due to behaviour analogous to a stretched string behaviour (3).

The SKM model can account for most of the BR characteristics. Distributions of \( x \) and \( \Psi \) (fig. 5) lead to broadening of \( \sim 1.5 \). This is due to distribution of \( \tau \) but not the activation energy (unlike DKG). Another important prediction is that \( \tau_0 (\geq 10^8/\nu) \) for GKM is two orders of magnitude higher than \( \tau_0 (\sim 10^2/\nu_0) \) for DKG. Unlike DKG, a well-defined stress level is not required to observe a decrease in \( \Delta \) when stress \( (\sigma > \frac{\mu b \sin \Psi}{x}) \) induces a pile-up of geometrical kinks.

2.3 The S-K relaxation. - S-K(\( \chi \)) relaxations where \( \chi \) 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 \( (\tau_{SK}) \) and strength \( (\Delta_{SK}) \) found using this model for a dilute concentration \( (C_d) \) of II's on the dislocations are given by:

\[
\tau_{SK} = \alpha kT L_0^2 C_d \exp \left( \frac{H_{SK}}{kT} \right); \quad \Delta_{SK} = B \Lambda L_0^2
\]

and the relaxation enthalpy \( (H_{SK}) \) by:

\[
H_{SK} = H^B + H^M
\]

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 \( \alpha, B \) are numerical constants. When the cores are saturated, \( H_{SK} \) reduces to \( H^M \). Recently, Verdini and Bacci (13) have developed a modification of the model which

Fig. 6 : Models of the S-K relaxation : a) string model. b) double-kink generation in the presence of core-mobile impurity interstitials.
takes into account the dissipation associated with bow-out between the II's. They show that \(\Delta_{SK}\) 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:

\[
H_{SK} = 2H_k + H'^M\tag{9}
\]

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

\[
H_{SK} = 2H_k + H'^M - \left(\mu \sigma b^6/2\pi\right)^{1/2}\tag{10}
\]

At the stress levels normally involved eqns |9| and |10| become practically identical. However, the last term does give an explicit dependence on \(\sigma\) (and hence \(\sigma_i\)) 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 (1379) 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 \(\Lambda\) 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.
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 $\mathcal{L}$ and is of the form $\mathcal{L}^r$ with $1 < r < 2$.

3.1.2 Peak temperature.- The peak temperature varies slightly with the microstructural state and is caused essentially by a dependence of $\tau$ on $\mathcal{L}$ ($\propto \mathcal{L}^s$ 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. $\tau$ 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 $I_F$ curves for slightly deformed Al in fig. 9 (18).
3.1.4 Non-Linear Effects.- When $\sigma_1$ is significant (after large plastic deformation), $\sigma_a$ has little effect (fig. 10). However, for weak $\sigma_1$

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; $n = 4$, $T_R = 246$ K; $n = 5$, $T_R = 285$ K; $n = 6$, $T_R = 316$ K). The short arrow indicates the peak temperature of $\Theta_2$, while the long arrow indicates the temperature of the maximum of $\delta_n - \delta_{n+1}$ (18).

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) $\varepsilon_m = 4 \times 10^{-7}$, (2) $15 \times 10^{-6}$, (3) $6 \times 10^{-6}$, (4) $5 \times 10^{-5}$. c) Internal friction in gold cold-worked 62% and annealed 60 min at 165°C. (1) $\varepsilon_m = 3 \times 10^{-7}$, (2) $1.5 \times 10^{-6}$, (3) $6 \times 10^{-6}$, (4) $1.5 \times 10^{-5}$ (19).
important non-linear effects are observed (Fig. 10): the relaxation strength depends strongly on $\sigma_a$ (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 $\sigma_a$. 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 $\sigma_s$ ($\sigma_m = 5 \times 10^{-6}$); 1, $\sigma_s = 0$; 2, $\sigma_s = 8 \times 10^{-6} \mu$; 3, $\sigma_s = 2 \times 10^{-5} \mu$; 4, $\sigma_s = 5 \times 10^{-5} \mu$; 5, $\sigma_s = 10^{-4} \mu$.

$B_1^*$ 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 (1's) give rise to $B_1$ and screws (0's) give rise to $B_2$, since the Peierls energy for 0's is higher than that for 1'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 Leiner (21) support an earlier hypothesis of Thompson and Holmes (22), i.e. $B_1$ is associated with 90° and 30° dislocations lying along <112> direction and $B_2$ is associated with 60° and screw dislocations lying along <110> direction. The 0'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 \( a_2 \) 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 role of stack-
ing 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 hy-
pothesis 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 experimen-
tal 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 \( \sigma_p \) ob-
tained from the relaxation enthalpy of the BR \( (10^{-3} - 10^{-4}) \mu \) is considerably grea-
ter 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
in response to the bias stress (25). In the tempe-
rate range of the BR, dislocation segments sur-
mount the Peierls barriers and the attenuation in-
creases with bias stress (fig. 12). Some different
results were reported re-
cently (26) but remain to
be confirmed. Secondly, a thermally activated microdeformation stage (3,27) associ-
ated with the BR has been observed in Al, Au and Mg (fig. 13) together with a va-
ration in the anelastic limit (27). A quantitative analysis of these results yields
an activation energy of \( \sim 0.1 \) eV and a volume of \( \sim 10^3 \) 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 \( 10^{-4} \) \( \mu \).
3.2 B.C.C. Metals.- The intrinsic dislocations relaxations in the b.c.c.'s are of special interest because the mobilities of \( \downarrow \)'s and \( \Theta \)'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 \( \downarrow \) and a peak at a much higher temperature associated with DKG \( \Theta \). Moreover, geometrical kinks on \( \Theta \)'s are expected to be rather abrupt (effectively high second order Peierls barriers \((30)\)) so that GKM \( \Theta \) should fall into a temperature range close to that for DKG \( \downarrow \). As in the case of the f.c.c.'s, GKM \( \downarrow \) is expected to occur at very low temperatures \((< 4 \text{ K})\) and manifest itself as a modulus defect at the lowest temperatures investigated. These expectations seem to be realized in Mo, W and \( \alpha \)-Fe as typified by the IFS of a deformed single crystal of Mo \((31,32)\) in fig. 14. In contrast, in V, Nb and Ta (also \( \alpha \)-Fe + H) the IFS is characterized by a larger number of peaks. Typical 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 II's.
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.

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).

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 $\alpha$ by most workers (35,36). However, a peak labelled $\delta$, at even lower temperatures, is observed in some samples (35,37). As suggested by Bruner (38), the Rome school (35,37) has argued that $\alpha$ requires the presence of both H and dislocations and that $\delta$ 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 $\delta$ nor any indication of dislocation movement below $\alpha$, leading them to attribute $\alpha$ to intrinsic dislocation processes. Nevertheless, over the last few years, evidence that $\alpha$ 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 $\delta$ in samples with lower II (particularly O) contents was offered. More recently, work on UHP Nb (31, 44, 45) has demonstrated that $\alpha$ involves H and that the major component can be attributed to an S-K(H) relaxation (43). In the highest purity samples ($C_H < 1$ 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 $\alpha$ ($\approx$ 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 $\delta$. In an attempt to do this, Maul (45) repeated measurement on sam-
samples 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 $\delta$ (35).

It is clear that in doped Nb there is a peak $\sim 30$ K which is equivalent to $\delta$. 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 ($\delta$) and 70 K peaks as having been demonstrated conclusively. Since $\delta$ is observed only in the presence of significant quantities of O (or N), we suggest that $\delta$ is predominantly GKM (0) and that DKG (1), 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 $\delta$ are more consistent with GKM (0) than DKG (1) (section 3.2.4); iii) results on thoroughly outgassed V (40) appear to exhibit both

Fig. 16: Internal friction of a single crystal of Nb after a thorough degassing ($C_H < 1$ 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 $C_H = 20-30$ at. p.p.m; 3, results of Mazzolai and Nuovo (35, 45).
peaks in the same spectrum, while the higher temperature peak is absent in samples outgassed to a lesser extent; iv) $\alpha$ (attributed to DKG (1) in $\alpha$-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 $\alpha$-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 $\alpha$) 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 $H^B(\downarrow) + H^M(\downarrow)$ and $2H_k(\downarrow) + H^M(\downarrow)$. From studies of peak $\alpha$ in Fe, $2H_k(\downarrow)$ is in the range 0.04 to 0.06 eV (48, 50, 51), while the most probable value of $H^B(\downarrow)$ is 0.61 eV (52). Whether we take $H^M = H^M(\downarrow)$ (following Seeger (4)) or $H^M = H^M(\downarrow)/2$ (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 $2H_k(\downarrow) + H^M(\downarrow)$ as pointed out by Kê et al. (49). We suggest that this very unsatisfactory situation arises because $2H_k(\downarrow)$ 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 $\downarrow$'s and softened $\Theta$'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 $\alpha'(0)$ and $\alpha$ processes; the high temperature peak is attributed to the $\gamma$ process. In the following, we refer to them simply as $\alpha'/\alpha$ and $\gamma$ respectively. Evidence for the presence of at least two discrete relaxations in $\alpha'/\alpha$ 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 $\alpha'/\alpha$ 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 $2\times10^{-8}$s; for Mo (3, 32), W (3) and $\alpha$-Fe (51, 46), the
equivalent ranges are 0.04 to 0.29 eV and $1.4 \times 10^{-14}$ to $5 \times 10^{-10}$ s. The first set, i.e. the old \( \delta \) peaks, seems to be consistent with GKM (\( \Theta \)); 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 \( \delta \) peaks.

**TABLE 1**

CLASSIFICATION SCHEME FOR THE IF PEAKS IN DEFORMED BCC METALS.

<table>
<thead>
<tr>
<th>PROCESS</th>
<th>OLD LABEL(S)</th>
<th>NEW LABEL</th>
</tr>
</thead>
<tbody>
<tr>
<td>INTRINSIC DISLOCATION</td>
<td>GKM (( \perp ))</td>
<td>( \alpha' )</td>
</tr>
<tr>
<td>LOW TEMP.</td>
<td>GKM (( \Theta ))</td>
<td>( \alpha'_2 )</td>
</tr>
<tr>
<td></td>
<td>DKG (( \perp ))</td>
<td>( \alpha_i )</td>
</tr>
<tr>
<td></td>
<td>( \delta ) (V, Nb, Ta)</td>
<td>( \alpha_i / \alpha )</td>
</tr>
<tr>
<td></td>
<td>57 K (V)</td>
<td>( i = 1, 2, 3 )</td>
</tr>
<tr>
<td></td>
<td>70 K (Nb)</td>
<td></td>
</tr>
<tr>
<td>INTRINSIC DISLOCATION</td>
<td>DKG (( \Theta ))</td>
<td>( \gamma, \gamma_1, \gamma_2 )</td>
</tr>
<tr>
<td>HIGH TEMP.</td>
<td></td>
<td>( \gamma (\perp) )</td>
</tr>
<tr>
<td>INTRINSIC INTERACTION</td>
<td>( \beta_i ), ( i = 1, 2, 3 )</td>
<td>( \beta_i (\perp) )</td>
</tr>
<tr>
<td>EXTRINSIC INTERACTION</td>
<td>GKM</td>
<td>( S-K' (X), S-K' (X,\perp) )</td>
</tr>
<tr>
<td>CORE-MOBILITY</td>
<td>DKG</td>
<td>( S-K (X), S-K (X,\Theta) )</td>
</tr>
<tr>
<td></td>
<td>( \alpha ) (V, Nb, Ta)</td>
<td>( X = H, O, N )</td>
</tr>
<tr>
<td></td>
<td>C.W.P.</td>
<td>( X = H, O, N )</td>
</tr>
<tr>
<td></td>
<td>S-K</td>
<td>( X = H, O, N )</td>
</tr>
<tr>
<td>EXTRINSIC INTERACTION</td>
<td>( \beta_i )</td>
<td>( \beta_\alpha (X), \beta_\gamma (X) )</td>
</tr>
<tr>
<td>IMMOBILE PINS</td>
<td>( p_i )</td>
<td></td>
</tr>
</tbody>
</table>

Relaxation parameters so far obtained for the \( \gamma \) peaks in Nb (56), Ta (33), Mo (32), W (3), \( \alpha \)-Fe (46, 57, 58) are all entirely consistent with DKG (\( \Theta \)). The \( \gamma \) peak narrowness and instability is taken into account by a model proposed by Astié (58). In this model, the \( \Theta \)'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 \( \gamma \) peak disappears. This model gives a good description of the \( \gamma \) behaviour in b.c.c.'s.
b) Conditions of Plastic deformation.- Studies of the amount and temperature of plastic deformation on the $\alpha'/\alpha$ and $\gamma$ peaks, particularly in Mo (32,59), W (77) and $\alpha$-Fe (58, 60) indicate that a major component of $\alpha'/\alpha$ is due to $\perp$'s and that $\gamma$ is due to $\Theta$'s. In $\alpha$-Fe the entire $\alpha'/\alpha$ complex is attributed to DKG (1) (60), which suggests that the apparent structure of $\alpha'/\alpha$ is due to the presence of several types of $\perp$'s with nearly equal mobilities.

c) Effects of Impurities.- In $\alpha$-Fe, the interaction of $\perp$'s (other than H) with $\perp$'s gives rise to $\beta_\alpha$ (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 $\beta_2$ by the author is attributed to the interaction of dislocations with unspecified impurities. In Nb (62) and Ta (fig. 15), the two components $\gamma_1$ and $\gamma_2$ that are candidates for DKG (9) behave differently in the presence of $\perp$'s. It has been reported that in Nb $\gamma_1$ requires the presence of O (62), whereas in Ta, $\gamma_2$ requires the presence of O (33). In Fe, $\beta_\gamma$ (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 $\gamma$ is a symmetric single peak (see fig. 14 for Mo). In substitutional Fe alloys, Astié (58) has observed two peaks called $\gamma_I$ and $\gamma_{II}$ (fig. 18): $\gamma_I$ appears around 230 K and $\gamma_{II}$ at about 320 K. For dilute alloys, $\gamma_I$ and $\gamma_{II}$ appear together whereas for concentrated alloys only $\gamma_I$ appears. The behaviour of these two peaks is quite similar to the $\gamma$ 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).](image-url)
The $\gamma_I$ peak is associated to DKG on softened $\Theta'$s and the activation enthalpy measured from IF experiments is in good agreement with the values deduced from plasticity behaviour (58). The $\gamma_{II}$ peak is due to DKG on normal $\Theta'$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 $\alpha$-Fe reported by the Grenoble group (46, 60, 61). The evolution of the IFS after 20K 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 $\alpha'/\alpha$ and generates $\beta_1$. After annealing in stage III of the recovery of electrical resistivity, the dislocations which give rise to $\alpha'/\alpha$ are cleaned of their intrinsic point defects and $\alpha'/\alpha$ reappears. By following changes in the relaxation intensity and peak temperature of $\alpha'/\alpha$ and $\beta_1$ 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 $\gamma$ have not been studied in any detail. A second interaction peak, $\beta_{\gamma}$, is observed at temperatures just below $\gamma$ in $\alpha$-Fe (46). It is attributed to the dragging of a self-interstitial-type defect by geometrical kinks on $\Theta'$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 $\alpha$-Fe (<4K) (60). These results are generally attributed to $\alpha'(1)$ (i.e. $\text{GKM}(1)$). However, in Fe there is some evidence that $\text{GKM}(\Theta)$ may be involved as well (60).
f) Effects of Oscillation Amplitude and Bias Stresses.- In the Nb single crystals (42), $\alpha'/\alpha$ increases in intensity with increasing strain amplitude. In $\alpha$-Fe, strain amplitude dependence occurs at $5 \times 10^{-6}$ even at 4K and increases with increasing temperature (61,63). Recent studies of the strain amplitude dependence associated with $\alpha'/\alpha$ and $\gamma$ in Mo single crystals (32) yield activation volumes of 30-150 $\text{b}^3$ for the major component of $\alpha'/\alpha$ and $\sim 50 \text{ b}^3$ for $\gamma$. These results are entirely consistent with DKG (L) and DKG (O) respectively.

Increasing strain amplitude causes the intensity of $\gamma$ 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 $\alpha'/\alpha$ (see fig. 21) and $\gamma$ (58,65) have been reported recently. The results obtained using a torsion pendulum technique can be summarized as follows:

1) the activation volume associated with $\alpha$ is in the range 10 to 100 $\text{b}^3$. 

---

Fig. 20: Evolution of the $\alpha'/\alpha + \beta_1$ complex in deformed $\alpha$-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 $\alpha'/\alpha$ and $\beta_1$ 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 $\alpha'/\alpha$ stage: $1 \sigma_s = 3 \times 10^{-5} \mu; 2 \sigma_s = 6 \times 10^{-5} \mu$ (64).
in keeping with DKG (1); ii) a substructure corresponding to \( \alpha' \) is observed but has not been analysed in detail; iii) the recovery stage \( \alpha_r \) (see fig. 20) (b)) in the absence of stress is always at a lower temperature than stage \( \alpha \); iv) a microdeformation stage is also associated with \( B_\alpha \) but its characteristics are such that it is not amenable to the same type of analysis used in the case of stage \( \alpha \); and v) the stage associated with \( \gamma \) (58,65) can be correlated with the zero-point-drift of the pendulum in conventional studies of the IF peak \( \gamma \) and the results confirm that \( \Theta \)'s are responsible for both the microdeformation stage and peak \( \gamma \). Furthermore, the model proposed by Astié et al. (65) (see 3.2.4. a) can also describe the \( \gamma \) microdeformation in b.c.c.'s.

### 3.2.5 Summary.

The IF peaks \( \alpha'/\alpha \) and \( \gamma \) (at \( \approx 1 \) Hz) are illustrated schematically (together with the S-K (H) peaks) in fig. 22. In Mo, W and \( \alpha\text{-Fe} \), \( \alpha'/\alpha \) contains a major component due to DKG (1), whereas in V, Nb and Ta containing significant amounts of both \( H \) and \( O \) (or \( N \)) it seems that DKG (1) is pinned-out and \( \alpha'/\alpha \) (previously \( \delta \)) is predominantly due to GKM (\( \Theta \)). However, in UHP Nb a major component of \( \alpha'/\alpha \) is attributed to DKG (1). In V, Nb, Ta and \( \alpha\text{-Fe} \) there is a good evidence that \( \alpha'(1) \) occurs below the lowest temperatures yet investigated.

**Fig. 22:** Schematic diagram showing the relative positions and shaped of the \( \alpha'/\alpha \) and \( \gamma \) 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 \( \alpha'/\alpha \) and \( \gamma \); 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 \( \alpha'/\alpha \) and \( \gamma \) taken together contain the analogue of the BR in the f.c.c.'s.

**Acknowledgements.** The authors wish to thank Prof. H. Schultz, P. Groh, C. Minier, M. Koiwa and Dr. C. Esnouf, J. Lauzier, L.B. Magalas who kindly made results of their groups' investigations available prior to publication.
References

(3) G. FANTOZZI, C. ESNOUF, W. BENOIT and I. RITCHIE, Prog. Mat. Science to be published.
J. SANJUAN, L. NO, G. FANTOZZI, C. ESNOUF and F. VANONI, this conference.
P. ASTIE, J.P. PEYRADE and P. GROH, Scripta Metall., to be published.