The double kink generation model, a good explanation for the Bordoni relaxation in high purity fcc metals

Autor(en): **Stadelmann, P. / Benoit, W.**

Objekttyp: **Article**

Zeitschrift: **Helvetica Physica Acta**

Band (Jahr): **52 (1979)**

Heft 5-6

PDF erstellt am: **25.09.2024**

Persistenter Link: https://doi.org/10.5169/seals-115046

Helvetica Physica Acta, Vol. 52 (1979), Birkhäuser Verlag, Basel

The double kink generation model, a good explanation for the Bordoni relaxation in high purity fcc metals[1])

P. Stadelmann[2]) and **W. Benoit**

Laboratoire de génie atomique, Swiss Federal Institute of Technology, 33, av. de Cour, CH-1007 Lausanne, Switzerland

(10. XII. 1979)

Abstract. The behaviour of the Bordoni relaxation in high purity fcc metals after plastic deformation and during recrystallization is compared with the theoretical predictions of the Double Kink Generation model. This model is developed here using both simplified analytical calculus (two well-model) and a complete computation. A very good agreement is found between experiments and theory. In particular, it is possible, using the DKG model, to follow the recovery of Al samples during heatings below the recrystallization temperature.

1. Introduction

As a consequence of the periodicity of the crystalline state the free energy of a dislocation, lying in a close-packed direction of the crystal, must be a periodic function of its position in the lattice. This variation of the free energy is known as the Peierls potential. In order to push the dislocation from a minimum of the free energy to the next it is necessary to apply to the dislocation a stress σ_p, the Peierls stress.

The jump between minima of the free energy is a thermally activated process. Consequently, a relaxation peak must be found using internal friction experiments. In fcc metals there is a relaxation peak which was attributed by A. Seeger [1] to this process. Using the experimental data a rather high Peierls stress is computed ($\sigma_p \approx 10^{-4}\,\mu$) with which many authors disagree. Moreover many aspects of the Bordoni relaxation have not been clearly explained (for instance disappearance of the relaxation during the recrystallization, unusual broadening of the peak).

In this paper a theoretical analysis of the Bordoni relaxation based upon the Double Kink Generation (DKG) model is developed and the effects of slight variations of its parameters are discussed and compared with the experimental results.

The analysis is performed in two ways: the first is a complete computed model which takes into account nearly all the parameters of the process and the other is

[1]) Research partially supported by the Swiss national Foundation, Subsidy no. 2. 932-0. 77.

[2]) Present address: Institut Interdépartemental de Métallurgie 33, av. de Cour, CH-1007 Lausanne, Switzerland.

based upon the simplified assumptions of the two-well model. The developed models confirm all experimental results.

Furthermore, the experimental study of the Bordoni relaxation in fcc metals of high purity (99.9999%) during recrystallization is able to show, on the one hand, that the disappearance of this relaxation in less pure fcc metals was generally due to impurities and, on the other hand, that in recrystallized fcc metals the effects of the residual stresses (internal stresses) can be clearly observed.

In the final section, it is shown that the use of the DKG theory allows one to model the annealing of Al 5 *N* samples and gives results which are in very good agreement with those of other methods (X-rays, microcalorimetry).

2. **Double kink generation model** (DKG)

In 1956, Seeger [1] proposed the following dislocation mechanism in order to explain the experimental results of Bordoni [2]. The free energy per unit length of a ⟨110⟩ dislocation, according to Peierls [3] and Nabarro [4] varies periodically as the dislocation slips in a {111} plane perpendicularly to a ⟨110⟩ direction. This energy can be written in the sinusoidal approximation as:

$$F_{\mathrm{Dis}} = F^0_{\mathrm{Dis}} + \sigma_p \frac{b \cdot d}{2\pi} (1 - \cos(2\pi y/d)) \tag{2.1}$$

where

$$\sigma_p = b^{-1} \left.\frac{dF_{\mathrm{Dis}}}{dy}\right)_{\max} \quad \text{is the Peierls stress}$$

b:Burgers vector of the dislocation

d:spacing between two dense directions

The Peierls stress σ_p is the stress which must be applied to the dislocation for this dislocation to make a jump over a Peierls hill at 0 K without internal stresses. This Peierls stress is a fundamental parameter for the plastic deformation of all crystalline solids and consequently it must be calculated very accurately.

When the crystal temperature is different from 0 K, the thermal fluctuations supply an amount of energy proportional to kT per jump and thus the additional energy which must be given to the dislocation is reduced. The mechanism for the jump over the Peierls hill is the nucleation of a double kink.

In Fig. 1 a ⟨110⟩ dislocation jumps over the maximum of F_{Dis} (Peierls hill) by making a double kink. The position 'b' is the unstable position corresponding to the activated state. When the internal stresses are too weak the kinks cannot move away from each other, and the dislocation returns to the 'a' position. On the other hand when there are sufficient stresses the position 'c' becomes the new equilibrium position and consequently the dislocation has performed a displacement of a distance d.

In real crystals the length of the dislocations is never infinite and an additional stress due to the increase of the dislocation length must be taken into account. This additional line tension limits the possible displacement of the dislocation to a few d (Fig. 2). Under the effect of the internal stresses σ_i the

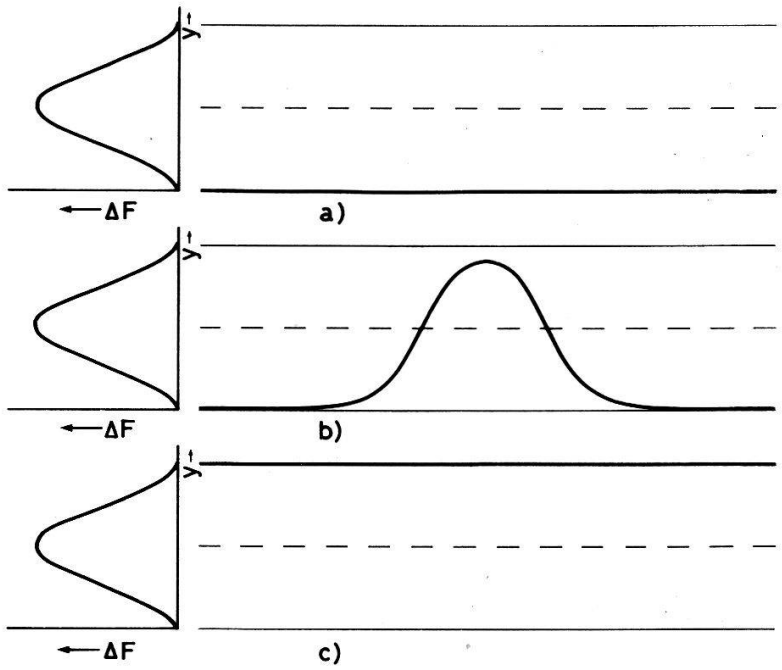

Figure 1
a) Stable equilibrium position. b) Unstable equilibrium position (activated state). c) Stable equilibrium position.

dislocations lie in the stable equilibrium position $1-S$. An amount of energy ΔF_N must be given to it to make a jump over the obstacle due to the Peierls stress and the line tension effect. ΔF_n is to be written as

$$\Delta F_N = \Delta F_0 + \Delta F_i \tag{2.2}$$

where ΔF_0 is the energy supplied by the thermal fluctuations (when the external stresses $\hat{\sigma}_a$ are zero), and ΔF_i is that supplied by the internal stresses. Thus, ΔF_0 is a function of σ_i (and $\hat{\sigma}_a$) and equals the activation energy. The activation area Δ_a

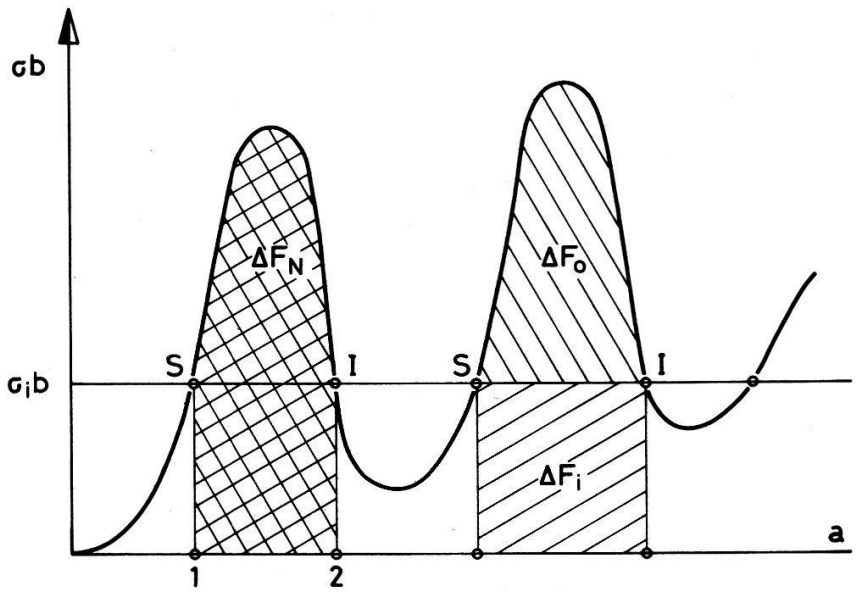

Figure 2
Effect of the line tension on the dislocation free energy.

is given by

$$\Delta_a = a_2 - a_1 = \Delta_a(\sigma_i) \tag{2.3}$$

During an activation event the applied stress (when $\neq 0$) furnishes an amount of work ΔW given by

$$\Delta W = \hat{\sigma}_a \cdot \Delta_a \cdot b \tag{2.4}$$

and consequently the thermal fluctuations have to supply only the following energy (free enthalpy):

$$\Delta G_N = \Delta F_0 - \Delta W \tag{2.5}$$

This free enthalpy is plotted in Fig. 3 as a function of the area swept by the dislocation during its movement.

The frequency of an activation event is in the present case given by the Boltzmann relation

$$\Gamma_{i,j} = \nu_G \exp(-\Delta G_{i,j}/kT) \tag{2.6}$$

where ν_G is the attempt frequency (constant in our formulation) and $i(j)$ defines the stable position of the dislocation before the jump (unstable position just after the activation).

In such a formulation the dislocation looks like a particle, with which one can attribute a probability n_i of being found in the ith stable position. This probability

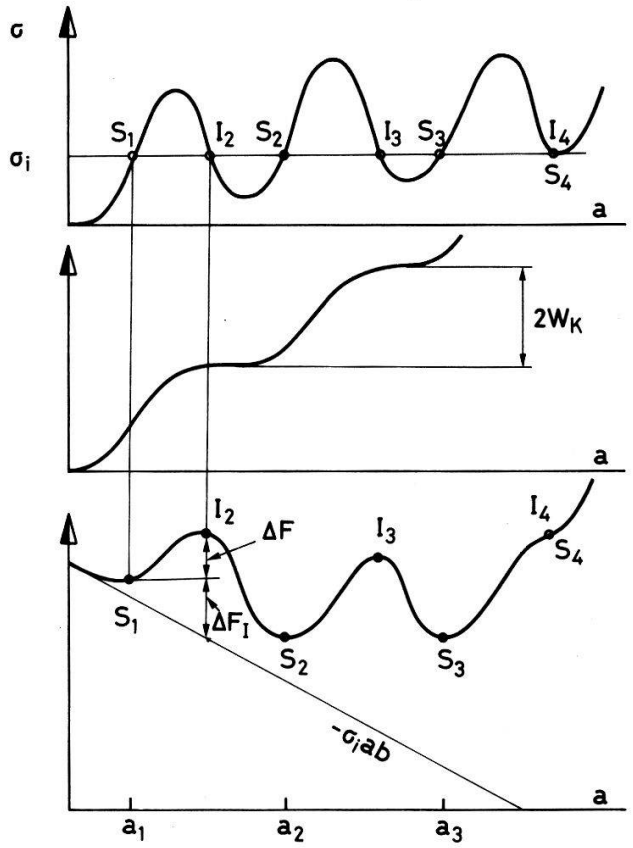

Figure 3
Schematic picture of the free enthalpy G of the dislocation showing the effect of σ_i (internal stresses).

is the solution of the following system of first order differential equations.

$$\frac{dn_i}{dt} = \Gamma_{i-1,i} \cdot n_{i-1} - (\Gamma_{i,i-1} + \Gamma_{i,i+1}) n_i + \Gamma_{i+1,i} n_{i+1} \tag{2.7}$$

Using this formulation the macroscopic anelastic deformation $\epsilon(t)$ is finally given by

$$\epsilon(t) = \Lambda_m \frac{b}{l} \sum_i n_i \cdot a_i + \epsilon(0) \tag{2.8}$$

where:

a_i: area swept by the dislocation in the ith minimum of G

Λ_m: density of the movable dislocations.

The experimental measurement of $\epsilon(t)$ can be performed in two ways. The first is the direct measurement of the microdeformation of the sample under a static external torsional stress. This technique is well adapted for the measurement of the activation area; but it is a rather difficult method and only a rough approximation of the nucleation energy can be obtained using it.

The second method is an indirect one. It consists of the measurement of the internal friction associated with the DK nucleation. Like all activated mechanism investigated through internal friction measurements the DKG is characterized by a relaxation peak.

This latter method is the oldest [5] and probably the most common. The internal friction measurement is performed by applying an external periodic stress to the sample (excitation stress) and by recording the loss of energy due to the DK nucleation. In this case the $\Gamma_{i,j}$ are written as

$$\Gamma_{i,j} = \nu_G \exp\{-\Delta G_{i,j} \pm b a_{i,j} \hat{\sigma}_a \sin(\omega t)\} \tag{2.9}$$

and the internal friction Q^{-1} is obtained using

$$Q^{-1} = \left(\frac{\mu}{2\pi\hat{\sigma}_a}\right) \oint \hat{\sigma}_a \sin(\omega t) \cdot \dot{\epsilon}\, dt \tag{2.10}$$

where μ is the associated shear modulus and the intergration is taken over one period.

Generally neither the microdeformation nor the internal friction can be computed analytically. Thus various approximations must be taken in order to give simple formulas. On the other hand a numerical computation can be very useful to fit the experimental results qualitatively. Such a computation will be examined in the following chapter. Simple formulas will be given in chapter 4 using a two-well model and compared with experimental datas collected during the recovery of deformed Al samples. A good agreement will be found.

3. Numerical computation and comparison with experiments

The assumptions of the computation are as follows:

- the dislocations are like elastic string each lying in a single Peierls minimum when the internal stresses are very weak

- the length of these dislocations is bounded by point defects from which thermal break-away cannot occur
- the dislocation line tension is isotropic and the sinusoïdal approximation for the Peierls potential is valid
- finally the dissociation of the dislocations is simulated by a decrease of the Peierls stress

Consequently the free enthalpy of the dislocation under the effect of a stress σ is given by the expression

$$G = F^0_{\text{Dis}} \int_0^l (ds - dx) + \int_0^l [\Delta F_{\text{Dis}}(y) - \Delta F_{\text{Dis}}(0)]\, ds - \int_0^l \sigma y\, dx \tag{3.1}$$

where the dislocation is lying along the x direction and $ds^2 = dx^2 + dy^2$ is the element length.

The boundary conditions are the following:

$$\begin{aligned} &y(0) = 0 \\ &y'(l/2) = 0 \quad \text{(symmetrical dislocation)} \end{aligned} \tag{3.2}$$

The extrema of G minima can be found using Euler's method [6]. This calculation was first performed by Arsenault [7] and will not be reported here.

The most important results of the numerical calculation of G minimum are that the free enthalpy ΔG_N is independent of the length of the dislocation; it is only a function of the stress σ. Secondly the nucleation area A_N is a weak function of the stress σ alone. Finally, this computation allows one to plot the 'theoretical equilibrium shapes' of the dislocation as a function of the swept area (Fig. 4).

It is rather surprising that the continuous model gives an unstable shape (curve 2, Fig. 4) which looks like a double kink.

Using the computed values for ΔG_N, ΔG_A, A_N and A_A for all the equilibrium position, one can compute the internal friction (formula 2.10). This computation is performed under the following assumptions:

- the dislocation density as a function of the length is given by the classical relation (3.3.):

$$\Lambda(l) = \Lambda_0 \cdot \frac{l}{l_0} \exp\left(-\frac{l}{l_0}\right) \tag{3.3}$$

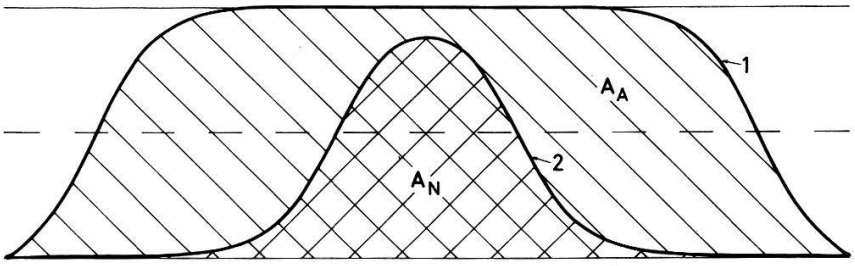

Figure 4
Equilibrium shapes of the dislocation as given by the continuous model (computed result).

- the internal stress is between 0 (recrystallized state) and $0.1\sigma_p$ (deformed state)
- the effect of impurities on the dislocation movement is always neglected (no solid solution hardening).

While it has been nearly impossible to simulate a real crystal with all its complexity, certain properties of the DKG can be investigated using this computational method and compared with experiments. Such properties are

- the effect of a variation of the measuring frequency on the internal friction associated with the Bordoni peak
- the effect of a decrease of the dislocation length
- the effect of the amplitude of the measuring stress without internal stresses or with important internal stresses ($\sigma_i \approx 0.1\ \sigma_p$).

Now compare the theoretical predictions with the experimental observations relating for each of these effects.

In Fig. 5 it is shown that an increase in the frequency of the measuring stress gives both an increase of the temperature and a broadening of the internal friction peak. Using the dependency of T_P with the frequency it is possible to calculate the relaxation energy associated with the peak and to compare it to the free enthalpy ΔG_N (Fig. 6).

This figure shows the logarithm of the measuring frequency, f_0, as a function of the inverse peak temperature T_P (Arrhenius plot). The relaxation energies $E_R(\sigma_i)$ which can be computed from the plot are in very good agreement with their corresponding free enthalpies ΔG_N.

As a consequence the best evaluation of the nucleation energy is experimentally obtained using the displacement of the peak under the effect of the variation of the measuring frequency [8, 9, 10].

The effect of a decreasing dislocation length, at constant internal stress, is reported in Fig. 7.

The Paré condition [11] is just fulfilled by the dislocations with a length of

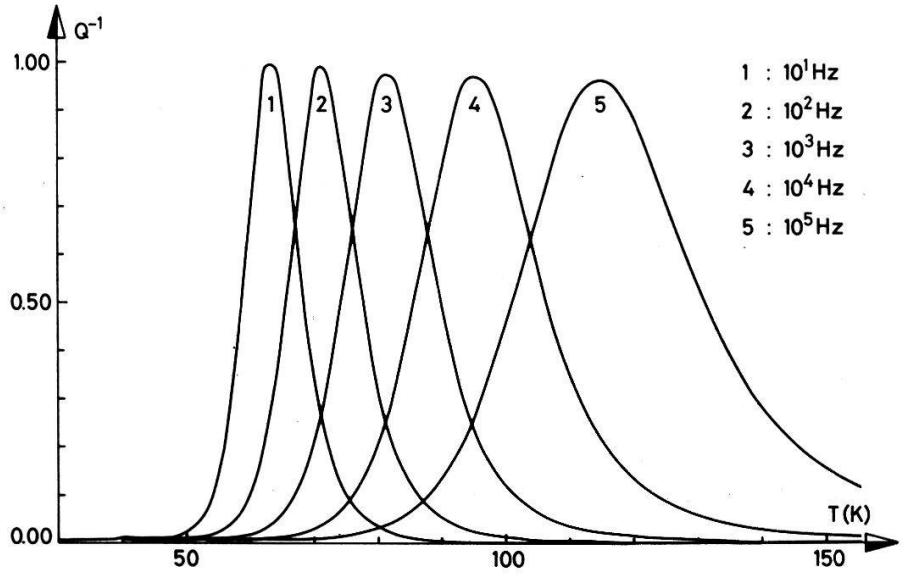

Figure 5
Effect on the peak of an increasing measuring frequency.

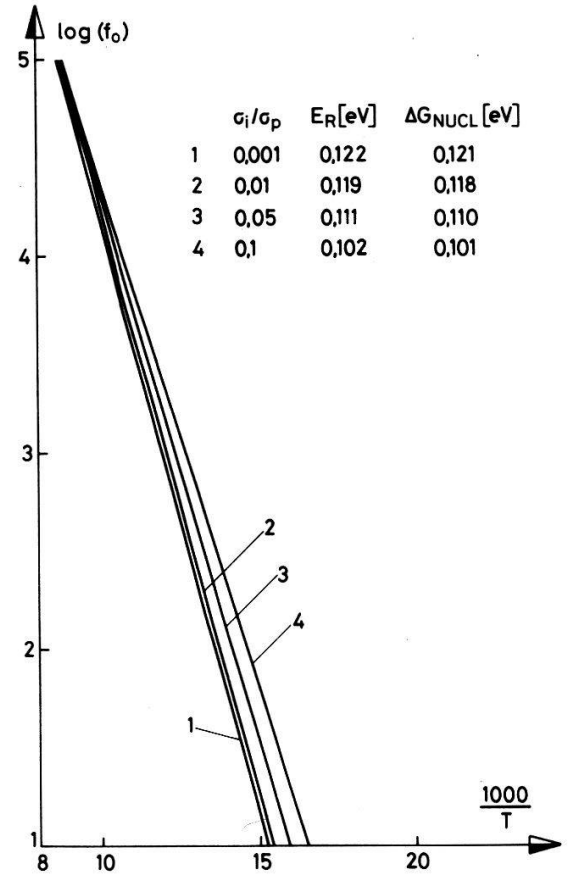

Figure 6
Arrhenius plot for the DKG and activations enthalpies as given by the slope of the curves (E_R) and the continuous model (ΔG_N).

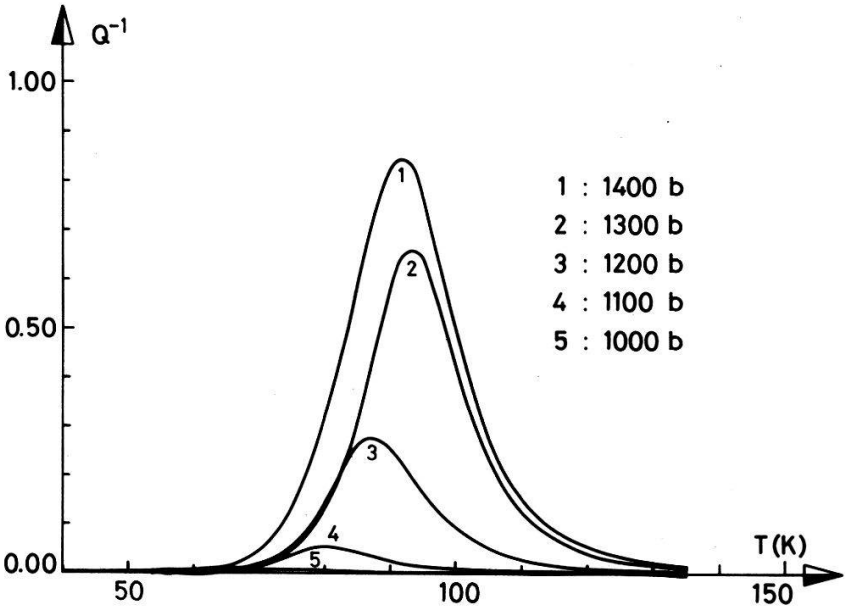

Figure 7
Effect of a decreasing dislocation length on the amplitude of the peak as given by the continuous model (computed result).

1400 b (curve 1) and is not fulfilled by shorter dislocations (curves 2–5) (the Paré condition will be explained in the next section).

From the Fig. 7 it can be seen that a decrease of the length is followed by a decrease of the peak temperature. This behaviour has been observed experimentally very often, perhaps the best known experiment being due to Thompson [12]. In this experiment Thompson using a sample of Cu, performed measurements of Q^{-1} as a function of irradiation dose (Fig. 8).

The decrease of the peak temperature with the increasing irradiation dose was explained by a preexponential factor ν_G dependent on the dislocation length ($\nu_G = \nu_D[b/l]$). In our case, without introducing such a dependency ($\nu_G =$ constant), the same behaviour can be very well explained by the broad variation of ΔG_A with the length l of the dislocations. The two-well model (next section) explains this behaviour.

The last important feature which was investigated using the computed DKG model is the effect of the measuring stress amplitude on the relaxation. In order to perform the computations it was assumed that a recrystallized sample can be characterized by zero internal stresses, and by rather long dislocations (about 1500 b measured in Ag 6 N after recrystallization [13]). This means that the purity of the sample must be high enough to eliminate pinning processes during and after the recrystallization. All these conditions are well met by very pure fcc samples (99.9999% [14]).

The theoretical predictions are shown in Fig. 9 for the case of a recrystallized specimen ($\sigma_i = 0$). One can see from this figure a fast decrease of the internal friction, and a peak temperature almost independent of the value of σ_a. Obviously, for small enough σ_a no relaxation will appear.

In Fig. 10 experimental curves are plotted. The sample contains less than 10 ppm Pt and presents a strong annealing at about 150 °C. This annealing is

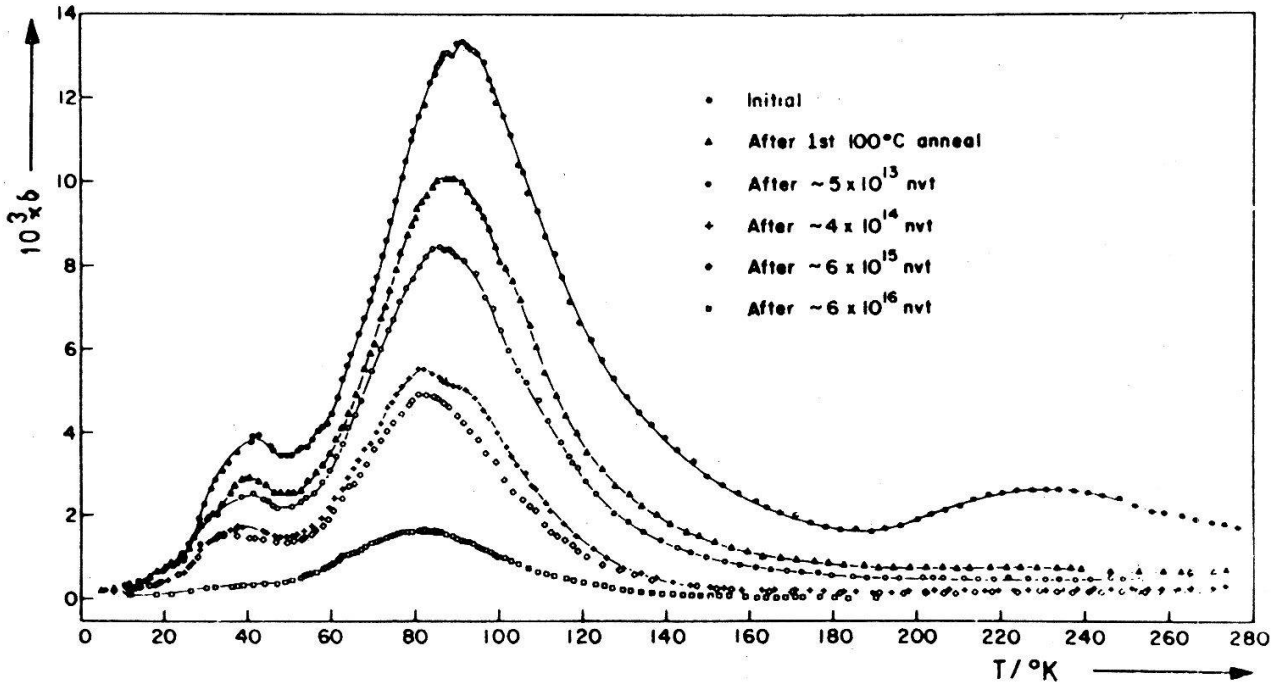

Figure 8
Thompson experiment with Cu sample of 99.999% purity.

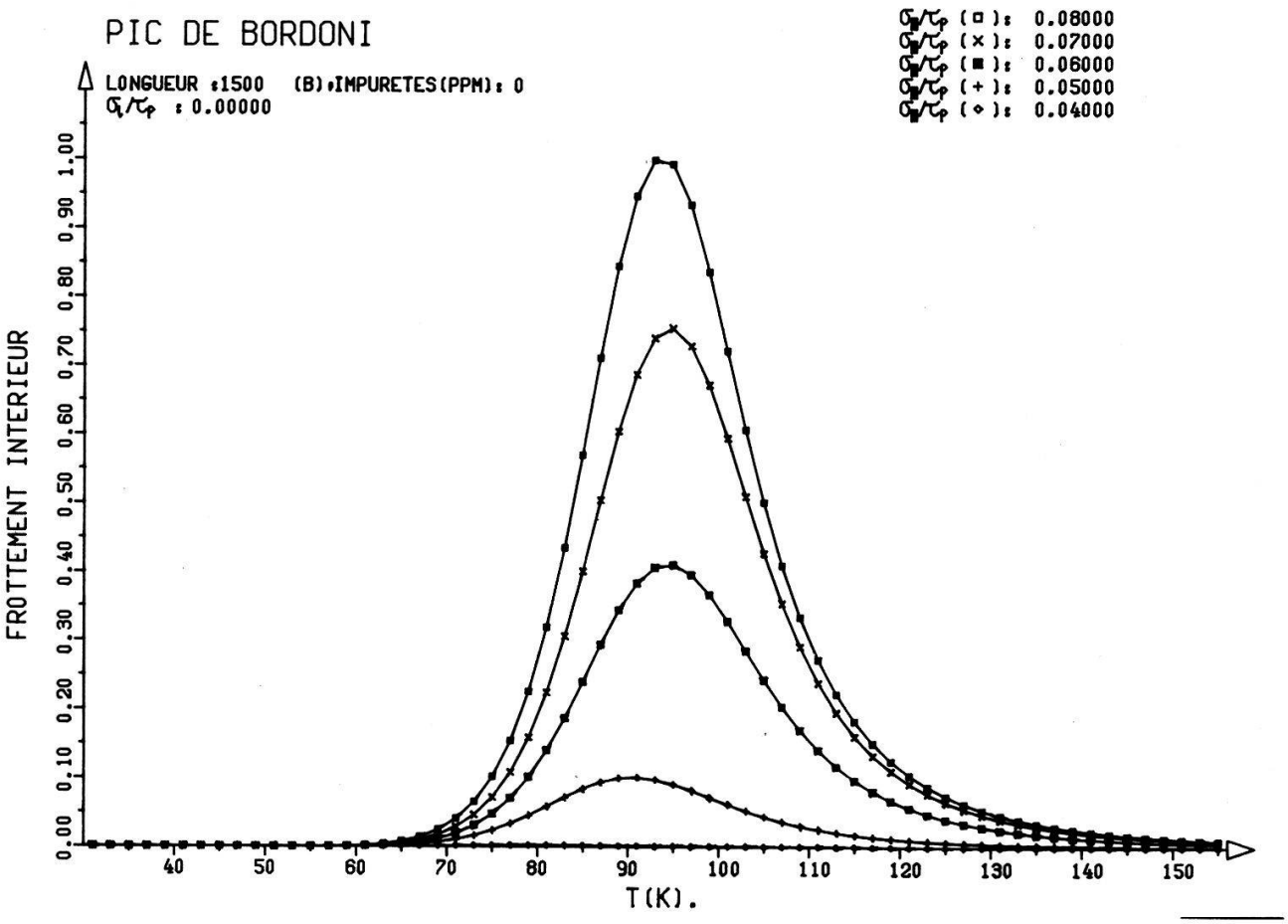

Figure 9
Effect of the measuring stress amplitude on the peak of a theoretical sample with $\sigma_i = 0$ (computed result).

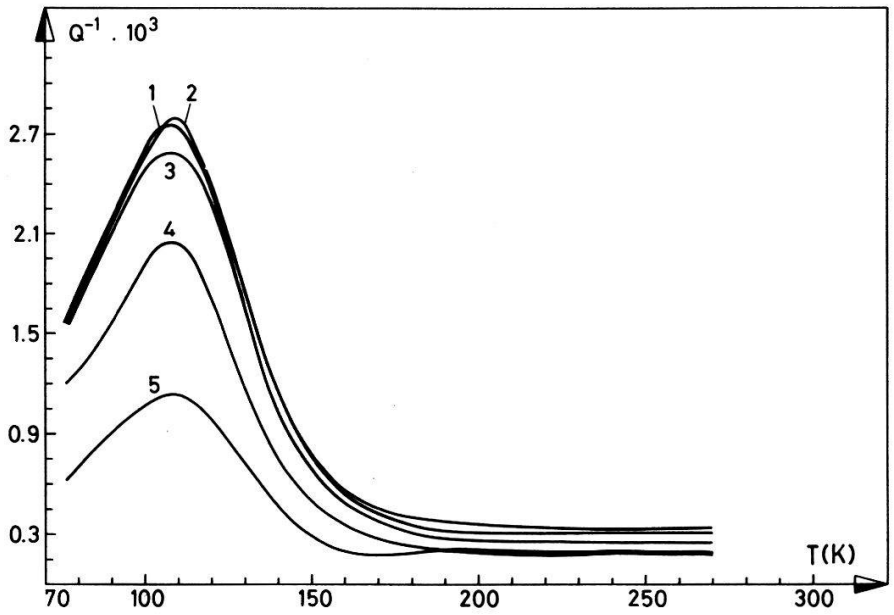

Figure 10
Same as Fig. 9, but experimental result with a sample of Au 99.999%.

characterized by the large effect of the measuring stress on the relaxation amplitude. This effect is not observed on the peak temperature.

The effect of the measuring stress was first observed by Jalanti [15]. It has now been reported for almost all the fcc metals [16, 17, 18] with purity higher or equal to 6 *N*.

In the case of heavily deformed samples ($\sigma_i \cong 0.1\ \sigma_p$) the theoretical internal friction does not vary with the amplitude of the measuring stress. This theoretical prediction is in good agreement with the experimental results.

Finally the most important limitation of the procedure comes from the great complexity of real samples. It is impossible to simulate real materials, even with the most time consuming computations. Only a good qualitative agreement can be found between theoretical and experimental results. With this limitation it is only the activation energy that can be measured and unambigously compared with the calculations.

4. Simplified analytical models

There are many simplified models for the DKG process [19, 20, 21]. Perhaps the most important one is due to Schlipf [21] which takes into account the principal behaviour of the DKG and that due to Chambers [22], the two-well model, which leads to very simple calculations and which will be developed here.

In the model due to Schlipf some very severe approximations about the DKG are made. Nevertheless, the analytical result gives a l^2-like dependence for the relaxation amplitude. This behaviour can not be extracted from the two-well model, and is certainly the principal limitation of the latter.

4.1. *Two-well model. Paré condition*

This model is schematized in Fig. 11. The transition rates $\overleftarrow{P}_t$ (backward) and $\vec{P}_t$ (forward) are given by the expressions:

$$\begin{aligned} \overleftarrow{P}_t &= \nu_G \exp(-\Delta G_A/kT) \\ \vec{P}_t &= \nu_G \exp(-\Delta G_N/kT) \end{aligned} \tag{4.1}$$

where ΔG_A and ΔG_N have the usual meaning.

Without stress applied to the dislocation the lateral movement of the kink is performed without energy expense. Thus in this case:

$$\Delta F_A = \Delta G_A(\hat{\sigma}_a = \sigma_i = 0) = 0$$

On the other hand the nucleation of a double kink forces the crystal to spend an amount of energy given by the expression:

$$\Delta F_N = \Delta G_N(\hat{\sigma}_a = \sigma_i = 0) \neq 0 \tag{4.2}$$

ΔG_N is due to an increase of the dislocation length and to the displacement of part of the dislocation to a region in the lattice of larger free energy.

In the case $\sigma_i = \hat{\sigma}_a = 0$, $\overleftarrow{P}_t \ll \vec{P}_t$ and the dislocation cannot move over the obstacle. In order to perform a jump $\overleftarrow{P}_t$ and $\vec{P}_t$ have to be approximatively equal. Thus:

$$\Delta G_A = \Delta G_N \cong \Delta F_0 \tag{4.3}$$

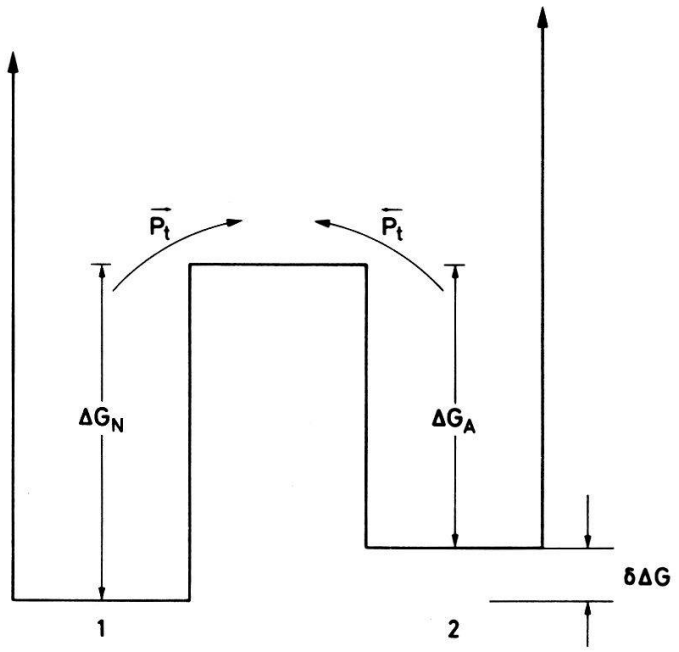

Figure 11
2 well-model of Chambers (1964).

This is the Paré condition. It can be expressed using the usual parameters in the formula

$$\Delta G_N = d \cdot b \cdot l \cdot \sigma \tag{4.4}$$

where the approximation $A_N = 0$ and $A_A = b \cdot l$ have been used.

We now introduce the probability n_2 for the dislocation to be in the second well. The time derivative is thus given

$$\dot{n}_2 = -\bar{P}_t \cdot n_2 + \vec{P}_t \cdot n_1 \tag{4.5}$$

where n_1 is the probability that the dislocation lies in the first well.

Using the notation $u \equiv (n_2 - \frac{1}{2}) \cdot d$, which means that the Paré condition is met at $t = 0$, it is possible to rewrite (4.5) as:

$$\dot{u} = -(\bar{P}_t + \vec{P}_t) \cdot u + \tfrac{1}{2}(\vec{P}_t - \bar{P}_t) \cdot d \tag{4.6}$$

Or:

$$B\dot{u} + Ku = \hat{\sigma}_a \cdot b\} \tag{4.7}$$

where

$B = (2\hat{\sigma}_a b/d)(\vec{P}_t - \bar{P}_t)^{-1}$: inverse of the dislocation mobility

$K = (2\hat{\sigma}_a b/d)(\vec{P}_t + \bar{P}_t)$: restoring force

It can be shown that the displacement of the dislocation under an oscillating applied stress $\sigma_a = \hat{\sigma}_a \exp\{i\omega t\}$ gives rise to a relaxation process. The relaxation time τ and the intensity Δ of the process are given by:

$$\begin{aligned} \tau &= B/K \\ \Delta &= \alpha \cdot \Lambda_m b^2 \mu'/K \end{aligned} \tag{4.8}$$

where α is an orientation coefficient and μ' is the shear modulus (unrelaxed state).

τ and Δ following (4.1) and (4.2) are now given by:

$$\begin{aligned} \tau(\sigma_i) &= \tfrac{1}{2}\nu_G^{-1} \exp\{\Delta F_0(\sigma_i)\}/kT_P \\ \Delta(l) &= \tfrac{1}{2}\alpha\Lambda_m\{\mu' b^3/kT_P\}\cdot b\cdot l \end{aligned} \tag{4.9}$$

Thus when the Paré condition is met the relaxation intensity Δ does not depend on the applied stress $\hat{\sigma}_a$ but depends linearly on the length of the dislocations. The first feature of Δ (that $\Delta \neq \Delta(\hat{\sigma}_a)$) is generally well verified in the case of the deformed sample for a large range of sample purities. On the other hand a l^2-like dependence is generally observed, but in certain cases the exponent can be slightly less than 2 [23].

Nevertheless the two-well model can be very helpful when the Paré condition is not completely fulfilled. This is generally the case when either the lengths of the dislocations are too short or the internal stresses are too weak. In both cases let us assume that the Paré condition is fulfilled by the dislocations with a length l_0. Thus for a shorter dislocation ($l < l_0$) we can assume that $n_2(0)$ is much less than $\frac{1}{2}$ and put $n_2(0) = 0$. Thus using (4.6) $u(t) = n_2 \cdot d$.

With these assumptions τ and Δ can be computed and the following results are obtained:

$$\begin{aligned} \tau(l) &= \tfrac{1}{2}\nu_G^{-1}\cdot\beta\cdot\exp\{\Delta F_0/kT_P\} \\ \Delta(l) &= \Delta(l_0)\cdot\beta\cdot(kT_P/2\Delta F_0) \end{aligned} \tag{4.10}$$

where

$$\beta = 1-(1-l/l_0)(\Delta F_0/2kT_P) \tag{4.11}$$

and

$$l \cong l_0 \quad \text{gives} \quad 1 - l/l_0 \ll (kT_P/\Delta F_0)$$

As a consequence, τ apparently decreases when l decreases, the relaxation energy being constant. The relaxation temperature which is generally given by the condition

$$\omega\tau = 1 \tag{4.12}$$

can now be written as

$$T_P^{-1} = (-k/\Delta F_0)\{\ln(\omega\tau_0) + \ln\beta\} \tag{4.13}$$

Thus T_P also decreases when the length of the dislocations decreases. On the other hand $\Delta(l)$ also decreases with l but much faster than the temperature (which decreases like $\ln(\beta)$).

This result can be used to explain why, even if ν_G is constant, the temperature of the peak can decrease during the pinning of the dislocations. Thus without assuming that ν_G is a function of the length of the dislocation the experimental results are well fitted by the DKG model.

4.2. *Applicability of the two-well model to the annealing of* A1 5 *N samples at liquid natrium and room temperature*

Samples of A1 5 *N* plastically deformed by rolling, show a very strong annealing stage when they are heated at temperatures near the recrystallization temperature. This annealing is performed in two stages and is observable using many methods. In particular X-ray diffraction, micro-hardness measurements, microphotography and microcalorimetry are the usually employed methods. But the most sensitive one is certainly the internal friction (associated with the modulus defect) measurement. Using X-ray diffraction and related other methods the following characteristics are found for these annealing stages:

- the first step presents the formation of cell walls inside the deformed grains. There are large energy releases, which are generally attributed to a decrease of the total dislocation density Λ_t [24, 25].
- the second step cannot be observed by either optical microscopy or electron microscopy. There is only a very weak energy release, and X-ray diffraction is also unable to detect it.

Using all these results it can be assumed that the first step is followed by a strong decrease of the internal stresses, which are related to the total dislocation density by:

$$\sigma_i = \eta \cdot \mu \cdot b \cdot \Lambda_t^{1/2} \quad \text{(Nabarro, [26])} \tag{4.14}$$

with $0.1 \leq \eta \leq 0.5$.

Introducing this formula into relation (3.3) and using the condition (4.12) lead us to:

$$T_{PB}(\Lambda_t) = T_{PB}(0) \cdot (1 - 2\eta \cdot \mu \cdot \Lambda_t^{1/2}/\sigma_p) \tag{4.15}$$

which gives the Bordoni peak temperature as a function of the total dislocation density Λ_t.

From formula (4.15) it can be deduced that a slight increase of the temperature of the peak may be observed during the first step of the annealing.

In order to compute this increase some evaluation of the change in dislocation densities is needed. The variation of the amplitude of the Bordoni relaxation can be used for these evaluations under the assumption that the Paré condition is fulfilled during the step. The latter assumption has been seen to be verified in recording the amplitude of the Bordoni relaxation as a function of the measuring stress [27]. It follows that relation (4.16) gives the peak amplitude as a function of the total dislocation density.

$$Q_{PB}^{-1} \alpha \Lambda_t \cdot l^2 \tag{4.16}$$

In order to use relationship (4.16) as a measure of the variation of the dislocation density the mean length l of the dislocations has to be a constant during the step. This was also verified [28].

The results of our analysis are that the variation of the ratio of the initial to final dislocation density, Λ_i/Λ_f, during the first step is about 2.4 for a specimen deformed at 77 K and 1.8 for a specimen deformed at 300 K. For comparison

purposes an analysis based upon the works by Fornerod [29] and Isoré [30] gives:

$\Lambda_i/\Lambda_f = 3.1$ deformed at 77 K

$\Lambda_i/\Lambda_f = 1.9$ deformed at 300 K

with no variation of the dislocation length during the first step of the annealing.

Thus there is a rather good agreement between the various methods. The DKG is also verified by the first annealing step in Al 5 *N*.

During the second step a large decrease of the Bordoni relaxation temperature is observed in both Figs. 12 and 13. These decreases are taken into account by the DKG model, but with two possibilities. The first one is to attribute the decrease to a very large increase in the dislocation density (relation 4.15), but this is meaningless. The second possibility is a change of the dislocation length and/or internal stresses, which as a consequence no longer fulfil the Paré condition. A large density decrease always leads to a large amount of released energy. Such a large amount is not observed by microcalorimetry. We did not observe a large change in the X-rays diffraction patterns, and furthermore we did not measure a strong dependence of the amplitude of the Bordoni relaxation with the measuring stress amplitude.

Consequently we can assume that the decrease of the relaxation amplitude and temperature is due to pronounced shortening of the dislocation length.

These assumptions allow us to compare the theoretical predictions of the DKG model with our experimental results.

Thus, using relations (4.10) and (4.11) we are able to compute the relative variation, $1 - l/l_0$, of the dislocation length. The first relation (4.10) uses the variation of the peak temperature and gives:

$$1 - l/l_0 = 2 \cdot k \cdot T_P(l_0)/\Delta F_0 \left\{ 1 - \exp \left\{ \frac{\Delta F_0}{k} \left[\frac{1}{T_P(l_0)} - \frac{1}{T_P(l)} \right] \right\} \right\} \tag{4.17}$$

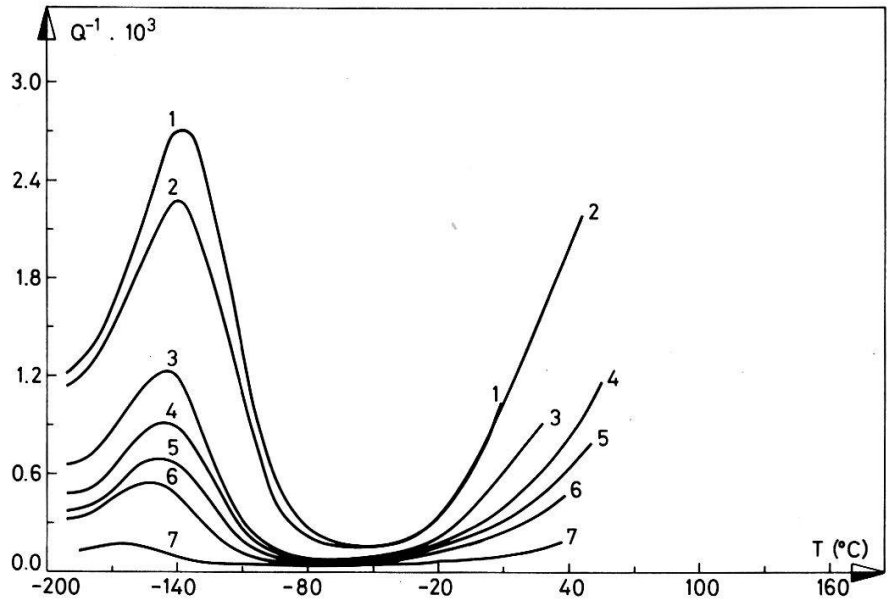

Figure 12
Al 5 *N* deformed 7% by rolling at 77 K.

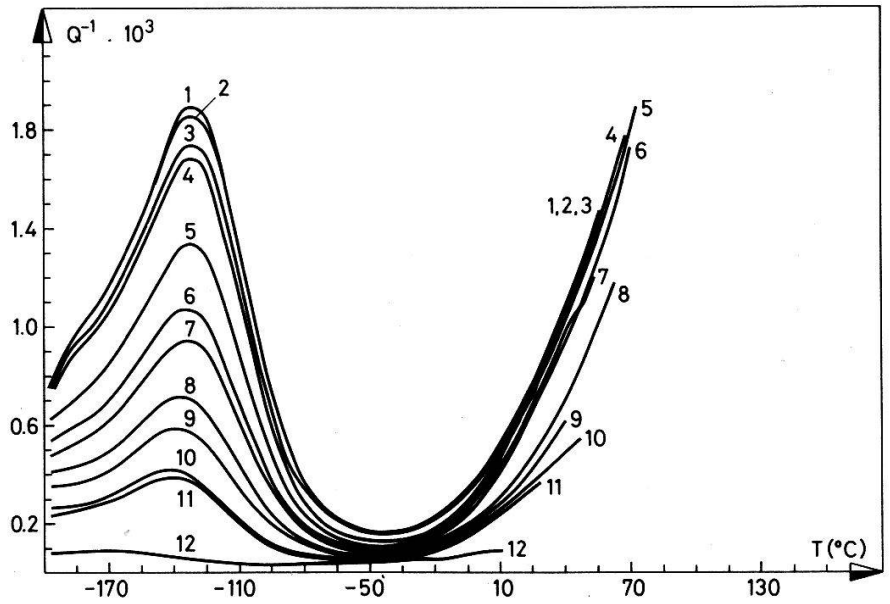

Figure 13
Al 5 *N* deformed 7% by rolling at 300 K.

The second one uses the variation of the peak amplitude and gives

$$1 - l/l_0 = 2 \cdot k \cdot T_P(l_0)/\Delta F_0\{1 - \Delta(l)/\Delta(l_0)\} \tag{4.18}$$

The relative variation l/l_0 is reported in Fig. 14, assuming that l_0 is the length of the dislocations just before the second step, at which time the Paré condition is equally just fulfilled.

There are good agreements between the results obtained using the T_P variation (14a) with those obtained using the Δ variation (14b). These agreements do not signify that the internal stresses are constant during the step, but they show clearly that the assumptions of the computation (on the Paré condition) were valid. However, the mechanism of the shortening of the dislocations cannot be

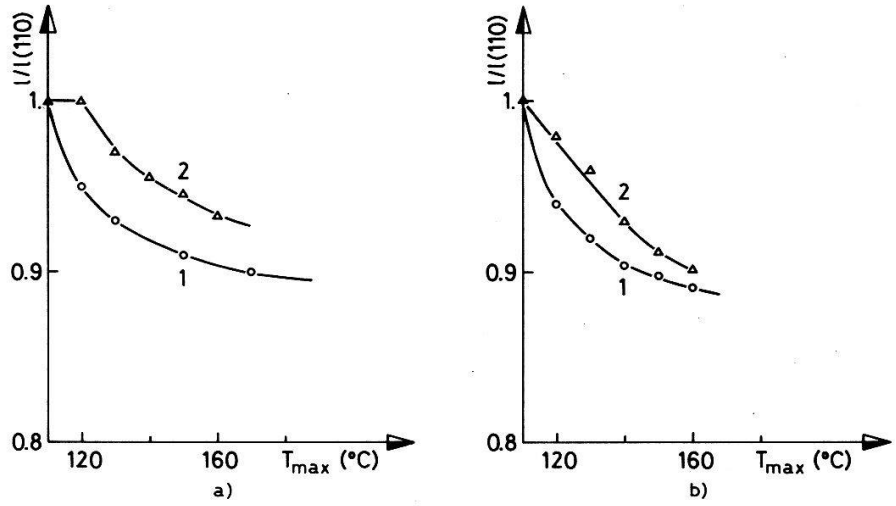

Figure 14
Part a) l/l_0 using the variation of the peak temperature. Part b) l/l_0 using the variation of the peak amplitude. 1 deformed 7% by rolling at 77 K. 2 deformed 7% by rolling at 300 K.

fully understood by studying only the change of the Bordoni relaxation. Nevertheless, these experiments show that the DKG model is verified and can explain all the observed behaviour of the Bordoni relaxation.

Naturally only by correlating various methods will one be able to clarify the physical origin of annealings. We cannot conclude that there is a dislocation pinning during the second step, but we can conclude that this mechanism looks like a pinning where seen through the DKG model.

5. Conclusion

The major results of this work are that the DKG model is able to explain almost all the behaviour of the Bordoni relaxation in fcc metals. The complete computed calculus, which is very useful in order to have a deep understanding of the DKG process, may be simplified in a two-well one without losing its main characteristics. The effects of impurities on the Bordoni relaxation can now also be understood and explained. Nevertheless it may be very instructive to perform new experiments using ultrasonic attenuation since the stress state inside the specimen is then much more finely resolved than in the case of vibrating samples. Perhaps using such a technique new behaviour will be obtained, which can confirm more closely the assumptions of the DKG model [31].

REFERENCES

[1] A. SEEGER, Phil. Mag. *1*, 651 (1956).
[2] P. G. BORDONI, Ric. Sci. *19*, 851 (1949).
[3] R. E. PEIERLS, Proc. Phys. Soc. *52*, 34 (1940).
[4] F. R. NABARRO, Z. S. BASINSKI and D. B. HOLT, Adv. in Phys. *13*, 193 (1964).
[5] C. WERT and C. ZENER, Phys. Rev. *76*, 1169 (1949).
[6] I. M. GELFAUD and S. V. FOMIN, *Calculus of Variations*, R. A. SILVERMAN Ed., Prentice-Hall N.J. (1963).
[7] H. J. ARSENAULT, Acta Met. *12*, 547 (1964).
[8] P. A. GRANDCHAMP, J. de Phys. *32*, C2-201 (1972).
[9] A. SEEGER, J. de Phys. *32*, C2-194 (1972).
[10] W. BENOIT, G. FANTOZZI, C. ESNOUF and J. PENZ, Ann. Phys. Fr. *4*, 7 (1979).
[11] V. K. PARÉ, J. Appl. Phys. *3*, 332 (1961).
[12] D. O. THOMPSON, in *Reinstoffprobleme E. Rexer* Akademic Verlag, Berlin *3*, 597 (1967).
[13] V. ROSSIER *Graduate work* EPF-Lausanne (1976).
[14] P. A. STADELMANN and W. BENOIT Scripta Met. *11*, 645 (1977).
[15] T. JALANTI, *Thesis* EPF-Lausanne (1975).
[16] Ref. cited (14).
[17] D. WAGNER and F. STANGLER, in *Internal Friction and Ultrasonic Attenuation in Solids*, Ed. R. R. HASIGUTI, University of Tokyo Press (1978).
[18] O. MERCIER, private communication (1978).
[19] U. F. KOCKS, A. S. ARGON and M. F. ASHBY in *Progress in Materials Science*, vol. *19*, Pergamon Press N.J. (1975).
[20] V. K. PARÉ, *Thesis* Cornell University Ithaca N.Y. (1958).
[21] J. SCHLIPF and R. SCHINDLMAYR, Nuovo Cimento *33* B, 22 (1976).
[22] R. H. CHAMBERS, in *Physical Acoustics*, Ed., P. MASON Academic Press, N.Y. (1966).
[23] P. A. GRANDCHAMP *Thesis* EPF-Lausanne (1972).
[24] R. A. VANDERMEER and P. GORDON, in *Recovery and Recrystallization of Metals*, Ed., L. HIMMEL, Gordon and Breach, N.Y. (1963).
[25] E. G. PERRYMAN, Acta Met. *2*, 26 (1954).
[26] Ref. cited (4).

[27] P. A. STADELMANN *Thesis* EPF-Lausanne (1978).
[28] G. FANTOZZI, *Thesis* INSA-Lyon (1971).
[29] R. C. FORNEROD, Mém. Sci. Rev. Mét. *66*, 2 (1969).
[30] A. ISORÉ, W. BENOIT and P. STADELMANN, Phil. Mag. *34*, 811 (1976).
[31] G. GREMAUD, W. BENOIT and R. GOTTHARDT, 3rd European Conf. on Internal Friction and Ultras. *Attenuation in Solids*, Manchester, July 79 (1979).