Application of Internal Friction to Analysis of Plastic Behaviour of Crystals

A. A. Lebedev

► To cite this version:

A. A. Lebedev. Application of Internal Friction to Analysis of Plastic Behaviour of Crystals. Journal de Physique IV Proceedings, 1996, 06 (C8), pp.C8-255-C8-264. 10.1051/jp4:1996855 . jpa-00254663

HAL Id: jpa-00254663

<https://hal.science/jpa-00254663>

Submitted on 1 Jan 1996

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.

Application of Internal Friction to Analysis of Plastic Behaviour of Crystals

A.B. Lebedev

Solid State Physics Division, A.F. Ioffe Physico-Technical Institute, Russian Academy of Sciences, Polytekhnicheskaya 26, St. Petersburg 194021, Russia

Abstract. A relationship between dislocation internal friction, microplastic and macroplastic behaviour of crystals has been considered. Main attention has been paid to the shape of the dislocation hysteresis responsible for the amplitude-dependent internal friction (ADIF) and to a comparison between the macroyield stress σ_e and microyield stress σ_e evaluated from the ADIF. It is shown that the same functional form of the ADIF curve may be due to different types of the dislocation hysteresis loop. A proportionality (similarity law) between the temperature dependences $\sigma_e(T)$ and $\sigma_e(T)$ has been demonstrated. Deviations from the similarity law can be used for analysis of the low and high temperature anomalies of $\sigma_e(T)$. The thermal vibrations of atoms rather than the thermal fluctuations are concluded to be responsible for the yield stress temperature dependence.

1. INTRODUCTION

A dislocation approach to crystal plasticity was suggested by Taylor [1], Orowan [2] and Polyani [3] in 1934. It was noted by Orowan [4] that the idea of a linear imperfection of the crystal lattice (in fact, a lattice dislocation) was put forward by Prandtl [5,6] as early as 1913 to explain the elastic hysteresis - one of the manifestation of the internal friction (IF). Thus, the lattice dislocations were first introduced in the field of internal friction (see a review by Seeger and Schiller [7] for details) and then used in crystal plasticity. However, in spite of the common roots, a relationship between the macroscopic plastic deformation and dislocation internal friction is still a subject of discussion.

Normally, the dislocation internal friction increases with increasing the vibrational strain amplitude ε_0 , and is considered to consist of amplitude-dependent and amplitude-independent parts. Davidenkov [8] in 1938 theoretically showed that the amplitude-dependent internal friction (ADIF) may result from a nonlinear hysteresis due to microplastic deformation. He assumed the microplastic strain ε_d to be a power function of the stress σ

$$\varepsilon_d \propto \sigma^m, \quad (1)$$

and obtained a power law for the amplitude-dependent decrement δ_h

$$\delta_h \propto \sigma_0^n, \quad (2)$$

where $\sigma_0 = M\varepsilon_0$ is the stress amplitude, M is the elastic modulus and $n=m-1$. Read [9,10] two years latter experimentally demonstrated a strong influence of plastic deformation on the internal friction within a wide range of ε_0 and pointed out the role of dislocations in hysteretic ultrasonic damping. Förster and Breithfeld [11] were the first to measure internal friction directly during plastic deformation and to find a component of IF depending on the strain rate. Bordoni [12] showed that plastic deformation leads to a relaxation peak in the IF temperature spectra.

On the other hand, there is an obvious difference between the macroscopic plastic deformation and dislocation internal friction (which is in different characters of dislocation motion, different barriers for

dislocations to overcome, different types of dislocations involved, different role of dislocation multiplication, etc.); therefore, mechanisms controlling these phenomena are generally considered to be different. However, many efforts (see, for example, papers by Brown [13], Schwarz et al. [14], and Kardashev [15]) have been done to elucidate a linkage between the dislocation internal friction, microplastic and macroplastic deformation. This linkage manifests itself in the following experiments:

(i) *Amplitude-dependent internal friction.* From the ADIF data it is possible to evaluate the stress-strain response in microplastic region and to determine the microyield stress, which can be compared to the ordinary (macroscopic) yield stress.

(ii) *Bordoni relaxation.* Analysis of the Bordoni relaxation makes it possible to estimate the Peierls stress σ_p . For pure materials, σ_p should be close to the critical resolved shear stress at low temperatures (near $T=0$).

(iii) *Internal friction during plastic deformation.* These measurements can be used to obtain *in situ* information on the evolution of the dislocation structure and active slip systems and also for studying the influence of various factors (such as illumination, ultrasonic vibrations, electric field, etc.) on plastic behaviour.

Problems of the Bordoni relaxation have been discussed in many review papers and monographs (see, for example, [7,16]); the most recent comprehensive reviews have been published by Fantozzi *et al.* [17] and Ritchi and Fantozzi [18]. Internal friction during plastic deformation have been recently reviewed by Lebedev [19,20]. The present paper deals with the amplitude-dependent internal friction.

Three peculiarities of the ADIF frequently observed in experiments at the early stages of the decrement amplitude dependence are important for the relation of the ADIF to micro- and macroplasticity.

(I) A proportionality between the amplitude-dependent decrement δ_h and modulus defect $\Delta M/M$:

$$\delta_h = r (\Delta M/M), \quad (3)$$

where r is a coefficient of the order of unity [9,10,21-33].

(II) A power law (2) for the decrement and modulus defect when n normally lies between 1 and 4 [9,10,21-39].

(III) A separation, $\delta_h = f_1(T)f_2(\sigma_0)$, of variables in the amplitude-temperature dependence of the decrement [21, 23-28, 30-32, 37, 39]. Taking into account this separation, Eqn.(2) can be rewritten as

$$\delta_h = f_1(T) \sigma_0^n, \quad (4)$$

2. NONLINEAR DISLOCATION HYSTERESIS AND MICROPLASTICITY

Baker [40] in 1962 suggested a simple algorithm to evaluate average dislocation velocities from the $\delta_h(\sigma_0)$ curves and, in fact, showed an empirical way to obtain the reversible microplastic (or dislocation) strain from the ADIF. He used Eqn.(3) and an approximation for the modulus defect (e.g. [16])

$$\Delta M/M \approx \varepsilon_d/\varepsilon_0, \quad (5)$$

then

$$\varepsilon_d \approx (\Delta M/M) \varepsilon_0 = \delta_h \varepsilon_0 / r. \quad (6)$$

Under steady state vibrations, the strain rate $\dot{\varepsilon}_d$ is proportional to the strain ε_d : $\dot{\varepsilon}_d = 4f\varepsilon_d$, where f is the frequency of vibrations and ε_d is, actually, the dislocation strain corresponding to the amplitude value of the elastic strain ε_0 (in this section, $\varepsilon_d(\varepsilon_0)$ will be denoted as ε_{dm}). Then, the dislocation velocity V can be expressed using the well known Orowan relation $\dot{\varepsilon}_d = \rho b V$ [4], where ρ is the dislocation density and b is the Burgers vector. It should be pointed out that Eqn. (6) is independent of the type of hysteresis loop. Let us consider the effect of the loop shape on the accuracy of the dislocation strain and velocity evaluated from the ADIF.

According to the classification given by Asano [41], there are two general models of nonlinear dislocation hysteresis responsible for the ADIF: the well known Granato-Lücke breakaway model [42], according to which a dislocation overcomes the same row of point defects during each half-cycle of vibration, and the frictional model, according to which dislocations move through various obstacles and this motion can be phenomenologically described by some frictional force. Asano derived general integral

equations between the $\varepsilon_d(\sigma)$ response and the $\delta_h(\sigma_0)$ dependence for both types of hysteresis loop. The frictional symmetrical loop equation

$$\delta_h(\sigma_0) = \frac{M}{\sigma_0^2} \left[2\sigma_0 \varepsilon_d(2\sigma_0) - 2 \int_0^{2\sigma_0} \varepsilon_d(\sigma) d\sigma \right] \quad (7)$$

has the solution (equation (25) in [41]), which allows one to evaluate the $\varepsilon_d(\sigma)$ response from the measured $\delta_h(\sigma_0)$ curve [41,43]. Nishino and Asano [43] used this solution to give an analytical expression for the dislocation velocity through the $\delta_h(\sigma_0)$ curve without any assumption on the modulus defect. In the particular case of Eqn.(2), the solution of Eqn.(7) yields [41] the Davidenkov loop.

However, the Davidenkov loop is not the only frictional symmetrical loop, satisfying Eqn.(7) in the case of the power function (2). One of the main features of the Davidenkov loop (the dashed line in Fig.1) is the presence of a restoring force. The solid line in Fig.1 is the loop with no restoring force (NRF). The NRF-type loops were, in particular, applied for the ADIF computer simulations [44,45]. In this loop, $\varepsilon_d(\sigma)$ remains constant (equal to ε_{dm} for decreasing and $-\varepsilon_{dm}$ for increasing the stress) until the stress reverses its sign. If we assume nonlinear branches of the NRF loop to obey Eqn.(1), then Eqn.(7) yields Eqn.(2) for the decrement (obviously, the NRF loop should be the same for both the absence of any restoring force and for a restoring force, which is in equilibrium with a pinning force at $\varepsilon_d(\sigma) = |\varepsilon_{dm}|$). One can define a parameter R , characterizing efficiency of the restoring force, in such a way that

$$\begin{aligned} \varepsilon_d &= -\varepsilon_{dm}, \text{ when } -R\sigma_0 > \sigma \geq -\sigma_0 \text{ and } \varepsilon_d = \varepsilon_d(\sigma), \text{ when } \sigma \geq R\sigma_0, \text{ for increasing } \sigma; \\ \varepsilon_d &= \varepsilon_{dm}, \text{ when } R\sigma_0 < \sigma \leq \sigma_0 \text{ and } \varepsilon_d = \varepsilon_d(\sigma), \text{ when } \sigma \leq R\sigma_0, \text{ for decreasing } \sigma. \end{aligned}$$

Then, if $\varepsilon_d(\sigma)$ obeys the power function (1), it can easily be obtained from Eqn.(7) that

$$\delta_h = 2M(1+R)^m \sigma_0^{m-1} (m-R)/(m+1). \quad (8)$$

If $R=1$, Eqn.(8) gives the decrement for the Davidenkov loop; if $R=0$, it gives δ_h for pure NRF loop. The broken line on Fig.1 is the loop for $R=0.5$. Theoretically, R may be varied from 1 to -1. The smaller is R , the weaker is the restoring force (i.e. the stronger is the pinning force). At $R=-1$, the loop transfers to a box, which gives amplitude independence of the internal friction.

It is clear from the above that Eqn.(7) has no unique solution; in other words, the same functional type of the $\delta_h(\sigma_0)$ curves may be due to different kinds of the hysteresis loop and, hence, to different microplastic strain-stress responses. Thus, the amplitude-dependent decrement only is not enough to evaluate the absolute values of the dislocation strain and dislocation velocity.

As has been shown by Lebedev [46,47], the ratio r from Eqn.(3) is sensitive to the type of the hysteresis loop. For the case of the power function (2), exact formulae $r(n)$ have been derived for the Granato-Lücke

Figure 1. Schematic representation of the frictional loops, giving Eqn.(8) for the decrement, at various R but for the same m (for this particular case, $m=3$). All the loops are normalized with respect to ε_d .

breakaway model and two frictional (Davidenkov and NRF loops) models (majority of the available experimental data are closer to the Davidenkov and NRF loops [47]). From the results of Refs.[46,47] for the case of Eqn.(2), it is easy to conclude that:

(1) the dislocation strain evaluated from the ADIF with the help of Eqn.(6) differs from the true ε_d value for both the Davidenkov and NRF loops by factors, which are functions of n only.

(2) The dislocation strain evaluated by the Asano algorithm (yielding the Davidenkov loop) also differs by a factor, depending only on n , from ε_d for the NRF loop.

(3) all the factors, within the practically interesting region ($n < 5$), are of the order of unity.

Thus, both of these methods (by Baker [40] and by Asano [41]) may give minor errors. Moreover, if the separation (4) takes place, the errors are the same for all temperatures where (4) is valid.

Fig.2 shows the $\delta_h(\sigma_0)$ curves measured at 100 kHz by Ivanov *et al.* [48] within the temperature range 6.5 - 295 K in Mg polycrystals

Figure 2. The amplitude-dependent decrement δ_h for increasing the strain amplitude ϵ_0 in commercial purity magnesium sample (data by Ivanov et al. [48]). The dashed line corresponds to $\epsilon_d = \text{constant}$.

Figure 3. The macroyield σ_c and microyield σ_e stresses in magnesium as a function of temperature. The points for σ_e , corresponding to $\epsilon_d = \text{constant}$ (see Fig.2), are plotted in relative units and compared to $\sigma_c(T)$ measured by Ivanov et al. [48].

Figure 4. Amplitude-dependent damping in Silicon Bronze at various temperatures on both increasing and decreasing the amplitude ϵ_0 [25,26]. The dashed line corresponds to $\epsilon_d \approx 2 \cdot 10^{-8}$. The solid lines are power functions $\delta_h \propto \epsilon_0^n$ with $n=1.2$.

with a crystallographic texture (the majority of grains had an orientation [0001]). Each of the curves corresponds to the first measurement on increasing the amplitude after annealing at room temperature for two days (this time was enough to recover the initial ADIF curve after ultrasonic excitation during the previous measurement). All the curves in Fig.2 exhibit a critical amplitude, at which a change in slope in the double logarithmic scale takes place. The curves on decreasing the amplitude showed higher values of the decrement for all investigated temperatures (they are not shown in Fig.2). Such amplitude hysteresis is generally attributed to an increase in the density of mobile dislocations under ultrasonic vibrations (due to multiplication or unlocking from the point defect atmospheres). The critical amplitude indicates the starting point for the multiplication (or unlocking). The dashed line in Fig.2 corresponds to Eqn.(6), i.e. it is a hyperbola $\varepsilon_d = \text{const}$. It is clear from Fig.2 that the critical amplitudes for various temperatures are on the line of constancy of ε_d and, hence, the dislocation multiplication (or unlocking from the atmospheres) does not take place until the dislocation strain reaches a certain value. In other words, the dislocation multiplication (unlocking) is determined by a critical level of the dislocation strain and depends on temperature only due to the fact that there is a temperature dependence of ε_d in the microplastic region (below the critical level); and the dislocation multiplication (unlocking) itself is an athermal process. The $\varepsilon_d(T)$ dependence follows from Eqns.(4)-(6) as

$$\varepsilon_d \propto f_1(T) \sigma^{n+1}. \quad (9)$$

Fig.3 shows the temperature dependence of the yield stress $\sigma_c(T)$, which is compared to the temperature dependence of the ultrasonic amplitude $\sigma_\varepsilon(T)$, provided a constant level of the dislocation strain below the critical level indicated by the dashed line on Fig.2. It is seen that both curves give a good coincidence (similarity) in relative units.

3. SIMILARITY LAW BETWEEN THE MICROYIELD AND MACROYIELD STRESSES

Lebedev and Kustov [48,50] used Eqn.(6) and showed that for a large variety of crystals at low ($T < 0.3T_m$) temperatures, the temperature dependence $\sigma_\varepsilon(T)$ of the microyield stress (evaluated from the ADIF as a stress at a constant level of ε_d) is proportional to the temperature dependence $\sigma_c(T)$ of the ordinary (macroyield) stress. For the manifestation of this similarity law, the ADIF should be approximated by Eqns.(3) and (4). Various definitions of the microyield stress (at a constant level of the decrement δ_h or the hysteresis loop area) were discussed elsewhere [51], and only the stress, providing a constancy of the dislocation strain, was demonstrated to be proportional to the macroyield stress. The microyield stress can be obtained from the ADIF at the level of ε_d as small as 10^{-10} - 10^{-8} , when the dislocation strain is completely reversible (it is clear from the Section 2 that both Eqns.(6) and (7) can be applied with essentially the same results). The absolute value of σ_ε can be one-two orders of magnitude smaller than σ_c .

This section gives some examples of the similarity law and shows application of the law to analysis of micro- and macroplasticity of crystals.

3.1. Experimental observations

Fig.4 shows the $\delta_h(\sigma_0)$ curves for polycrystalline silicon bronze measured by Pilecki et al. [24] at various temperatures. The solid lines correspond to the power law (2) with $n=1.2$. The dashed line is a hyperbola $\varepsilon_d \approx 2 \times 10^{-8}$. The points of intersection of the $\varepsilon_d \approx 2 \times 10^{-8}$ and $\delta_h(\sigma_0)$ curves give the temperature dependence of the microyield stress $\sigma_\varepsilon(T)$. Comparison between the $\sigma_\varepsilon(T)$ and $\sigma_c(T)$ measurements demonstrates (Fig.5) that the similarity law in the silicon bronze holds well [25].

The similarity law in zone refined *Mo* single crystal [37] is valid at $T < 180$ K (Fig.6). A deviation from the similarity is observed at $T > 180$ K (when the point defects become mobile and the ADIF is not a single-valued function of the amplitude any more). The deviation is due to the process of immobilization of dislocations by point defects, which is much more effective for the microyield than for macroyield stress. This type of deviation is very well pronounced in impure *Al* [26,27].

It is well known that sometimes the temperature dependence of the yield stress in crystals, instead of a normal increase of the stress with decreasing T , exhibits maxima (or flattening) at low ($T < 40$ K)

Figure 5. The macroyield and microyield stresses in polycrystalline Silicon Bronze (data by Lebedev and Pilecki [25]). Relative change in both σ_c and σ_ϵ scales is the same.

Figure 6. The similarity law in Mo single crystal [37]. τ_c is the critical resolved shear stress, τ_ϵ is the microyield shear stress evaluated from the ADIF.

Figure 7. The macroyield τ_c (after Didenko and Pustovalov [54]) and microyield τ_ϵ (after Lebedev and Ivanov [27], Schwarz and Granato [55], and Kosugi and Kino [56]) shear stresses in aluminium single crystals.

Figure 8. Data by Goto et al. [38] for the microyield stress and by Feltham and Meakin [58] for the macroyield stress in Cu show that the similarity law with respect to the grain size is not valid.

temperatures. The nature of the low T anomaly is still under discussion. Several mechanisms have been suggested in the literature: (i) inertial or thermo-inertial overcoming of point defects [14,52]; (ii) quantum tunneling of dislocations through barriers [52]; (iii) local heating of slip bands [53]; (iv) changing the dislocation multiplication contribution to the yield stress [53].

Fig.7 gives an example of the anomaly of $\sigma_c(T)$ measured by Didenko and Pustovalov [54] in Al single crystal. The microyield stress, evaluated from the ADIF data by Lebedev and Ivanov [27], Schwarz and Granato [55], and Kosugi and Kino [56], shows no anomaly. Thus, the deviation from the similarity law allows one to identify the physical mechanisms of the low T anomaly. In particular, the results for pure Al, presented in Fig.7, exclude all the mechanisms, but (iv).

It is more difficult to explain the similarity law itself than its violation. At present, the similarity law is observed in single crystals of different lattice structure and various purity: *NaCl*, *NaF*, *LiF*, *MgO*, *W*, *Al* [49,50], *Mo* [37,49,50] and *Al-Si-Fe* [26,27] and in polycrystals: *Mg* [49,50], silicon bronze [25] and *Fe-Cr* alloys [57].

It should be noted that there is no similarity between the microyield and macroyield stresses as a function of the grain size (see Fig.8, where the data by Goto *et al.* [38] for σ_e are compared in the Hall-Petch plot with the data by Feltham and Meakin [58] for σ_c) and the impurity concentration [39,59]. From the ADIF data measured by Lebedev *et al.* [29] during the early stage of plastic flow of *NaF* single crystals, it is clear that σ_e decreases with deformation, in contrast to a slight increase in σ_c .

Nishino and Asano [60] and Nishino *et al.* [61], investigating the ADIF in *Al* thin films, found a proportionality between σ_e and σ_c as a function of film thickness t . Both stresses vary as $1/t$, which can be well explained by a pinning of dislocation segments at the film surface and film-substrate interface [61].

If there exists any physical reason for the similarity law with respect to temperature, it should be related to the origin of the yield stress temperature dependence.

3.2. Temperature dependence of the yield stress

The yield stress temperature dependence is generally considered [62-65] to be a consequence of the thermofluctuational nature of plastic deformation when the plastic strain rate can be express as

$$\dot{\varepsilon}_d = \dot{\varepsilon}_0 \exp[-H(\sigma^*)/kT], \quad (10)$$

where $\dot{\varepsilon}_0$ is a pre-exponential factor, $H(\sigma^*)$ is the stress-dependent activation enthalpy and $\sigma^* = \sigma - \sigma_i$ is the so-called effective stress, which is the difference between the applied and internal stresses. The latter is considered to be independent of temperature (with an accuracy of the elastic moduli temperature dependence). Then, $\sigma(T)$ can be obtained by solving Eqn.(10) under the condition $\dot{\varepsilon}_d = \text{const}$. In this case $\sigma(T)$ is mainly determined by the function $H(\sigma^*)$, which depends strongly on the specific mechanism of dislocation-barrier interaction.

In b.c.c. metals, screw dislocations are much less mobile than non-screw ones; therefore the macroplastic deformation is controlled by the Peierls barrier - screw dislocations interaction [66,67]. Existence of the ADIF in pure b.c.c. metals (see, e.g., [68,69]) can be attributed only to the interaction of edge (or mixed) dislocations with point defects.

In polycrystals, microdeformation (at $\varepsilon_d < 10^{-6}$) occurs inside a small amount of independent individual grains, whereas a collective shear propagation takes place during macrodeformation. Grain boundaries play minor role in the former case, but they are effective barriers for dislocations in the latter case [70-72].

Validity of the similarity law in polycrystals (Fig.5) and in pure b.c.c. metals (Fig.6), where the mechanisms of micro and macroplasticity are known to be different, allows us to conclude that the similarity law is not in agreement with Eqn.(10) [25,73].

Alternative approach to the description of $\sigma(T)$ is to account for the thermal vibrations of atoms rather than the thermal fluctuations. Dietze [74] and Khulman-Wilsdorf [75] applied this approach and calculated a reduction of the Peierls stress σ_p with increasing temperature. Leibfried [76] and Saul and Bauer [23] showed that the stress for overcoming point defects by a dislocation also depends on the mean-square amplitude of atomic vibration $\langle u^2 \rangle$. Nabarro [77] recently pointed out that the results obtained by Dietze [74] and Kuhlman-Wilsdorf [75] can approximately be expressed in the same form:

$$\sigma_p = \sigma_0 \exp [-A T], \quad (11)$$

where σ_0 is the stress at $T=0$ and A is a material-dependent factor, which is 10/3 times higher in Ref.[74] than in Ref.[75]. According to Wang [78], the data on $\sigma_c(T)$ for crystals, having low σ_p , agree better with the Kuhlman-Wilsdorf theory, whereas high- σ_p crystals are in consistent with the Dietze prediction.

In contrast to the thermofluctuational concept, the $\langle u^2 \rangle$ approach does not relate the temperature dependence $\sigma(T)$ to the strain rate. Independence of the strain rate (at a constant stress) and a proportionality of the strain and stress rates are features of the stress-induced plasticity [79-81]. Eqn.(9), which obeys both condition, describes, in fact, the stress-induced plasticity. If the temperature-dependent function in (9) has a form $f_i \propto \exp [A T]$, then the stress at $\dot{\epsilon}_d = \text{const}$ will obey Eqn.(11), frequently observed (see, for example, Fig.6) in experiments for microyield [26,27,37] and macroyield [82,83] stresses.

The $\langle u^2 \rangle$ approach may be a basis for the similarity law. A proportionality between the experimental $\langle u^2 \rangle$ values and the macroyield stress, reported recently by Butt and Chaudhry [83], confirms this assumption.

4. CONCLUSIONS

(1) Microplastic stress-strain response can be evaluated from the ADIF in absolute units only if there is information about the shape of the dislocation hysteresis loop, or on the amplitude-dependent modulus defect. There is one-to-one correspondence between the former and the latter.

(2) The similarity law $\sigma_s(T) \propto \sigma_c(T)$ can be explained by the temperature dependence of the stress-induced plasticity.

(3) Such temperature dependence is provided by taking into account the mean square amplitude of atomic vibrations (the function $f_i(T)$ in Eqn.11 plays a role of the Debye-Waller factor).

(4) Three components contribute to the yield stress: (i) σ_i is independent of temperature but depends on the dislocation structure, (ii) σ_A^* depends on temperature in accordance with the Arrhenius-type equation (9), which is responsible for the strain rate sensitivity, relaxation and creep, and (iii) σ_s^* depends on temperature due to $\langle u^2 \rangle$. The latter component gives the main contribution to the yield stress temperature dependence, which leads to the similarity law.

(5) Deviations from the similarity law may be used to identify the mechanism of the yield stress temperature dependence anomalies.

Acknowledgments

The author is grateful to Dr. M.M. Korsukova, Prof. O.V. Klyavin, Dr. B.K. Kardashev, Dr. S.B. Kustov, Prof. S.P. Nikanorov, and Dr. A.E. Romanov (Ioffe Phys.-Tech. Institute, St. Petersburg), Prof. A.V. Granato (University of Illinois, Urbana), Prof. S. Asano and Dr. Y. Nishino (Nagoya Institute of Technology) for useful discussions.

References

1. Taylor G.I., *Proc. Royal Soc.* **A145** (1934) 362-387.
2. Orowan E., *Z. Physik* **89** (1934) 634-659.
3. Polyani M., *Z. Physik* **89** (1934) 660-664.
4. Orowan E., *Proc. Phys. Soc. (London)* **52** (1940) 8-22.
5. Prandtl L., (1913) unpublished work cited by T. Kármán (*Enzyk. Math. Wiss.*, Bd. 4, S.767, 1913) (reference is taken from [4,6,7]).
6. Prandtl L., *Z. angew. Math. Mech.* **8** (1928) 85-106.
7. Seeger A., Schiller P., *Physical Acoustics*, Vol.3, Part A, Ed. by W.P. Mason (Academic Press, New York, 1966) pp. 361-495.
8. Davidenkov N.N., *Zh. Tekhn. Fiz.* **8** (1938) 483-499.

9. Read T.A., *Phys. Rev.* **58** (1940) 371-380.
10. Read T.A., *Trans. AIME* **143** (1941) 30-41.
11. Förster F., Breithfeld H., *Z. Metallkunde* **30** (1938) 343-345.
12. Bordoni P.G., *Ric. Sci.* **19** (1949) 851-862; *J. Acoust. Soc. Amer.* **26** (1954) 495-502.
13. Brown N., Microplasticity, Ed. by C.J. McMahon, Jr. (Wiley, New York, 1968) pp.45-73.
14. Schwarz R.B., Isaac R.D., Granato A.V., *Phys. Rev. Letters* **38** (1977) 554-557.
15. Kardashev B.K., *Fiz. Tverd. Tela* **19** (1977) 2490-2496 [*Sov. Phys. Sol. State* **19** (1977) 1459-1462].
16. Nowick A.S., Berry B.S., *Anelastic Relaxation in Crystalline Solids* (Academic Press, New York, 1972).
17. Fantozzi G., Esnouf C., Benoit W., Ritchie I.G., *Progr. Mat. Sci.* **27** (1982) 311-451.
18. Ritchie I.G., Fantozzi G., *Dislocation in Solids*, Vol.9, Ed. by F.R.N. Nabarro (North-Holland, Amsterdam, 1992) pp. 57-133.
19. Lebedev A.B., *Fiz. Tverd. Tela* **35** (1993) 2305-2341 [*Phys. Sol. State* **35** (1993) 1141-1159].
20. Lebedev A.B., *Mat. Sci. Forum* **210-213** (1996) 519-526.
21. Nowick A.S., *Phys. Rev.* **80** (1950) 249-257.
22. Takahashi S., *J. Phys. Soc. Japan* **11** (1956) 1253-1261.
23. Saul R.H., Bauer C.L., *J. Appl. Phys.* **39** (1968) 1469-1477.
24. Pilecki S., Ivanov V.I., Lebedev A.B., *Phys. Stat. Sol. (a)* **119** (1990) 87-92.
25. Lebedev A.B., Pilecki S., *Scripta Met. Mater.* **32** (1995) 173-178.
26. Lebedev A.B., *Fiz. Tverd. Tela* **34** (1992) 1889-1896 [*Sov. Phys. Sol. State* **34** (1992) 1008-1012].
27. Lebedev A.B., Ivanov V.I., *Mat. Sci. Forum* **119-121** (1993) 245-250.
28. Lebedev A.B., Nikanorov S.P. *Fiz. Tverd. Tela* **38** (1996) 839-850 [*Phys. Sol. State* **38** (1996) 465-471].
29. Lebedev A.B., Kustov S.B., Kardashev B.K., *Fiz. Tverd. Tela* **31** (1989) 62-68 [*Sov. Phys. Sol. State* **31** (1989) 34-38].
30. Kardashev B.K., Nikanorov S.P., Voinova O.A., *Fiz. Tverd. Tela* **16** (1974) 1068-1072 [*Sov. Phys. Sol. State* **16** (1974) 687-689].
31. Nikanorov S.P., Kardashev B.K., *Elasticity and dislocation anelasticity of crystals* (Nauka, Moscow, 1985) (in Russian).
32. Kardashev B.K., Kustov S.B., Lebedev A.B., Berezhkova G.V., Perstnev P.P., Appel F., Messerschmidt U., *Phys. Stat. Sol. (a)* **91** (1985) 79-87.
33. Lebedev A.B., Burenkov Yu.A., Golubenko T.I., *Fiz. Tverd. Tela* **35** (1993) 420-430 [*Phys. Sol. State* **35** (1993) 216-221].
34. Gelli D., *J. Appl. Phys.* **33** (1962) 1547-1550.
35. Kharitonov A.V., *Akust. Zh.* **9** (1963) 80-87 [*Sov. Phys.-Acoustics* **9** (1963) 63-69].
36. Gelli D., *Phys. Stat. Sol.* **12** (1965) 829-841.
37. Lebedev A.B., Kardashev B.K., Hofmann U., Kaufmann H.-J., Schulze D. *Cryst. Res. Technol.* **24** (1989) 1143-1149.
38. Goto H., Nishino Y., Asano S. *J. Japan Inst. Met.* **55** (1991) 848-852.
39. Nishino Y., Okada Y., Asano S. *Phys. Stat. Sol. (a)* **129** (1992) 409-419.
40. Baker G.S., *J. Appl. Phys.* **33** (1962) 1730-1732.
41. Asano S. *J. Phys. Soc. Japan* **29** (1970) 952-963.
42. Granato A.V., Lücke K., *J. Appl. Phys.* **27** (1956) 583-593.
43. Nishino Y., Asano S., *Phys. Stat. Sol. (a)* **151** (1995) 87-91.
44. Schwarz R.B., *Acta Metall.* **29** (1981) 311-323.
45. Belan V.I., Landau A.I., *Fiz. Met. Metalloved.* **65** (1988) 259-267 [*Phys. Met. Metallog.* **65(2)** (1988) 44-51].
46. Lebedev A.B., *Philos. Mag. A* **74** (1996) 137-150.
47. Lebedev A.B., Proc. ICIFUAS-11, Poitiers, France, July 7-11, 1996 (this issue).
48. Ivanov V.I., Lebedev A.B., Kardashev B.K., Nikanorov S.P., *Internal Friction and Fine Structure of Metals and Inorganic Materials*, Ed. by F.N. Tavazze (Nauka, Moscow, 1985) pp.102-105 (in Russian).
49. Lebedev A.B., Kustov S.B., *Fiz. Tverd. Tela* **29** (1987) 915-918 [*Sov. Phys. Sol. State* **29** (1987) 524-525].
50. Lebedev A.B., Kustov S.B. *Phys. Stat. Sol. (a)* **116** (1989) 645-656.
51. Lebedev A.B. *J. Alloys and Compounds* **211-212** (1994) 177-180.

52. Startsev V.I., Dislocations in Solids, Vol.6, Ed. by F.R.N. Nabarro (North-Holland, Amsterdam, 1983) pp.143-234.
53. Klyavin O.V. Physics of Plasticity of Crystals at Liquid Helium Temperatures (Nauka, Moscow, 1987) pp. 218-222 (in Russian).
54. Didenko D.A., Pustovalov V.V., *J. Low Temp. Phys.* **11** (1973) 65-76.
55. Schwarz R.B., Granato A.V., *Phys. Rev. Letters* **34** (1975) 1174-1177.
56. Kosugi T., Kino T., *J. Phys. Soc. Japan* **53** (1984) 3837-3842.
57. Golovin I.S. *Metal. and Mater. Trans. A* **25** (1994) 111-124.
58. Feltham P., Meakin J.D., *Philos. Mag.* **2** (1957) 105-112.
59. Morita M., Asano S., *J. Japan Inst. Met.* **57** (1993) 1006-1011.
60. Nishino Y., Asano S., Strength of Materials, Proc. 10th Int. Conf. (ICSMA-10), Sendai, Japan, August 21-26, 1994, Ed. by H. Oikawa, K. Maruyama, S. Takeuchi and M. Yamaguchi (Japan Inst. Met., Sendai, 1994) pp. 857-860.
61. Nishino Y., Tanahachi K., Asano S., *Mater. Trans. JIM* **36** (1995) 1476-1482.
62. Seeger A., *Philos. Mag.* **45** (1954) 771-773; *ibid.* **46** (1955) 1194-1217.
63. Fridel J. Dislocations (Pergamon Press, Oxford, 1964).
64. Kocks U.F., Argon A.S., Ashby M.F. Thermodynamics and Kinetics of Slip, *Progr. Mater. Sci.* **19** (Pergamon Press, Oxford, 1975).
65. Schoeck G. Dislocations in Solids, Vol.3, Ed. by F.R.N. Nabarro (North-Holland, Amsterdam, 1980) pp. 63-163.
66. Louchet F., Kubin L.P., Vesely D. *Philos. Mag. A* **39** (1979) 433-454.
67. Suzuki H. Dislocations in Solids, Vol.4, Ed. by F.R.N. Nabarro (North-Holland, Amsterdam, 1979) pp. 191-217.
68. Pal-Val P.P., Kaufmann H.-J., *Phys.Stat.Sol.(a)* **62** (1980) 569-575.
69. Kardashev B.K., Kaufmann H.-J., Lebedev A.B., Pal-Val P.P., *Phys.Stat.Sol.(a)* **104** (1987) K17-K20.
70. Suits J.C., Chalmers B. *Acta Met.* **9** (1961) 854-860.
71. Brentnall W.D., Rostoker W. *Acta Met.* **13** (1965) 187-193.
72. Dudarev E.F. Microplastic deformation and Yield Stress of Polycrystals (Izd. Tomskogo Universiteta, Tomsk, 1988) (in Russian).
73. Lebedev A.B. Strength of Materials, Proc. 10th Int. Conf. (ICSMA-10), Sendai, Japan, August 21-26, 1994, Ed. by H. Oikawa, K. Maruyama, S. Takeuchi and M. Yamaguchi (Japan Inst.Met., Sendai, 1994) pp. 195-198.
74. Dietze H.-D., *Z. Phys.* **132** (1952) 107-110.
75. Kuhlmann-Wilsdorf D. *Phys.Rev.* **120** (1960) 773-781.
76. Leibfrid G., Dislocation and Mechanical Properties of Crystals, Ed. by J.C. Fisher, W.G. Johnston, R. Thomson, and T. Vreeland Jr. (Wiley, New York, 1957) pp. 495-500.
77. Nabarro F.R.N., *Mater. Sci. Engng. A* **113** (1989) 315-326.
78. Wang J.N., *Mater. Sci. Engng. A* **206** (1996) 259-269.
79. Klyavin O.V., Likhachev V.A., Nikiforov A.V., Shvetsova V.A. *Vest. Leningrad Univ. Ser. Math., Mech., Astronom.* No.7 (1978) 75-83.
80. Likhachev V.A., Shvetsova V.A. *Problemy Prochnosti* No.11 (1984) 57-62 [*Strength of Materials* **16** (1984) 1570-1576].
81. Alden T.H., *Metall. Trans. A* **16** (1985) 375-392.
82. Butt M.Z., Felham P., Ghaury I.M., *J. Mater. Sci.* **21** (1986) 2664-2666.
83. Butt M.Z., Chaudhry S.A., *Phys. Rev. B* **47** (1993) 8418-8424.