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MECHANICS

Determining Characteristic Plastic-Relaxation Times Using Micro- and Nanocrystalline Nickel as an Example

E. N. Borodin[a], N. S. Selyutina[b], and Corresponding Member of the RAS Yu. V. Petrov[b, c]

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Abstract—Based on the concept of the incubation time of plastic deformation, an integral yield criterion is introduced and time effects of irreversible deformation are considered. The efficiency of the approach is demonstrated using micro- and nanocrystalline nickel as an example. The parameters of the phenomenological model are treated physically from the viewpoint of the behavior of the defect structure of the material, which is controlled by the dislocation sliding and grain-boundary slip mechanisms in a wide range of the rate of deformation.

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The classic approaches to determining the yield stress based on the Mises or Tresca criterion as the moment of attaining a threshold value of stresses in a material do not consider the majority of works devoted to the plastic deformation of metals. Although numerical studies have shown that the use of the classic approaches leads to errors in the case of dynamic deformation [1], in some cases, which are traditionally considered as quasi-static deformation, relaxation processes also occur during the deformation of a material. Among these cases are whisker crystals of pure metals [2], as well as numerous nanosized objects [3] and bulk nanostructured metals [4]. In all these cases, the role of the relaxation processes, which are responsible for inertness in the development of plastic flow, in the formation of the mechanical response of a material at the initial stages of the deformation process can be not only comparable with the role of static barriers, but also dominant. In this work, we consider an integral yield criterion [5–8] with the parameters of the barrier stress (static yield stress) and the characteristic plastic-relaxation time (incubation time). Based on the concept of the characteristic stress-relaxation time, which depends only on the state of the defect substructure of a material, the maximum stresses in a particular specimen of material can be determined at a specific strain rate and other external conditions of loading.

[a] *Ural Federal University, Mira str. 19, Ekaterinburg, 620002 Russia*
[b] *St. Petersburg State University, Universitetskii pr. 28, Peterhof, St. Petersburg, 198504 Russia*
[c] *Institute for Problems in Mechanical Engineering, Russian Academy of Sciences, Bol'shoi pr. 61, Vasil'evskii ostrov, St. Petersburg, 199178 Russia*
e-mail: elbor7@gmail.com; yp@yp1004.spb.edu

THE RELAXATION MODEL OF PLASTIC DEFORMATION

Let us assume that, in addition to elastic stresses, the rheological relation for the material subjected to dynamic deformation contains a plastic component, which is proportional to the strain rate. This leads formally to the integral criterion of the onset of plastic flow [9, 10], which can be interpreted as follows using the fading-memory concept [5, 6]:

$$\mathrm{Int}_p(t) \le 1, \quad \mathrm{Int}_p(t) = \frac{1}{\tau}\int_{t-\tau}^{t}\left(\frac{\Sigma(s)}{\sigma_y^0}\right)^{\alpha} ds. \qquad (1)$$

Here, $\Sigma(t)$ is the time dependence of the stresses; τ is the characteristic stress-relaxation time (incubation period); σ_y^0 is the static yield stress; and α is the stress-sensitivity factor. Under the assumption of elastic deformation $\Sigma(t) = 2G\varepsilon(t)$, this criterion makes it possible to calculate the moment of the onset of macroscopic flow t_*, which corresponds to the onset of equality in (1).

We consider the simplest version of the relaxation model of plasticity [9, 10]. Let the linear increase in the deformations in the specimen obey the law $\varepsilon(t) = \dot{\varepsilon}tH(t)$, where $H(t)$ is the Heaviside function. We introduce the dimensionless relaxation function $0 < \gamma(t) \le 1$ using the following condition:

$$\gamma(t) = \begin{cases} 1, & \mathrm{Int}_p(t) \le 1, \\ \left(\mathrm{Int}_p(t)\right)^{1/\alpha}, & \mathrm{Int}_p(t) > 1. \end{cases} \qquad (2)$$

Let us assume that the stress $\Sigma(t)$ averaged over the incubation period and exceeding the right-hand side of inequality in (1), "relaxes" in such a manner that at the moments of time $t \ge t_*$, which correspond to the

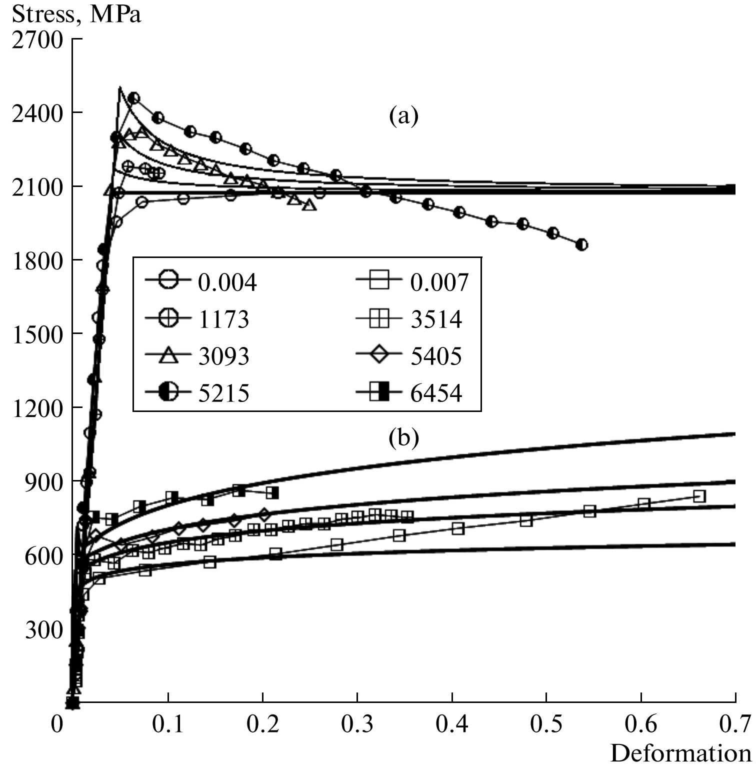

Fig. 1. Illustration of the phenomenon of sharp yield point for (a) nanocrystalline nickel without strain hardening and (b) microcrystalline nickel with strain hardening in a wide range of strain rate (s^{-1}) plotted using (1)–(4) the relaxation model of plasticity and experimental data [4].

plastic deformation of the material, the following condition is satisfied [10]:

$$\mathrm{Int}_p(t)\gamma(t) = 1. \tag{3}$$

The true stresses in the specimen being deformed at $t \geq t_*$ are determined using the following relation:

$$\sigma(t) = 2g(t)\varepsilon(t), \tag{4}$$

where $g(t) = G\gamma^{1-\beta}(t)$ and β is the scalar parameter ($0 \leq \beta \leq 1$), which controls the degree of strain hardening. The case $\beta = 0$ corresponds to the absence of strain hardening.

The lines without symbols in Fig. 1 show the stress–strain curves for monocrystalline (σ_y^0 = 438 MPa, $G = 76$ GPa, and the grain size is 48.44 μm) and nanocrystalline ($\sigma_y^0 = 2072$ MPa, $G = 25$ GPa, and the grain size is 17 nm) nickel plotted at various rates of deformation using model (1)–(4). The symbols correspond to the data from [4]. It can be seen that the calculated curves agree well with the experimental data at the initial stage of deformation and that a sharp yield point appears in the stress–strain curves with an increasing strain rate. The characteristic relaxation time for nickel increases nearly ten times when the grain size decreases from 48.44 μm (τ = 0.575 μs) to 17 nm (τ = 3.3 μs).

DEPENDENCE OF RELAXATION TIMES ON THE PARAMETERS OF THE DEFECT STRUCTURE OF A MATERIAL

According to [9, 10], the dependence of the characteristic relaxation time on the parameters of the defect structure of the material in the case of dislocation plasticity can be expressed as follows: $\tau_D = \dfrac{B_f}{Gb^2\rho_D}$, where B_f is the coefficient of phonon friction for dislocations [1]; b is the strength of dislocations; and ρ_D is the density of dislocations. In Fig. 1, the time of 0.575 μs for microcrystalline nickel [4] corresponds to the density of dislocations of $\sim 10^{11}$ m^{-2}, which is typical of the specimens of the metals used in the experiments.

With a decrease in the grain size, the dominant plastic deformation mechanism changes from dislocation sliding into grain-boundary slip [11]. These mechanisms are characterized by different values of the yield stress of the material [12] and different relaxation times. In the case of plasticity that follows the

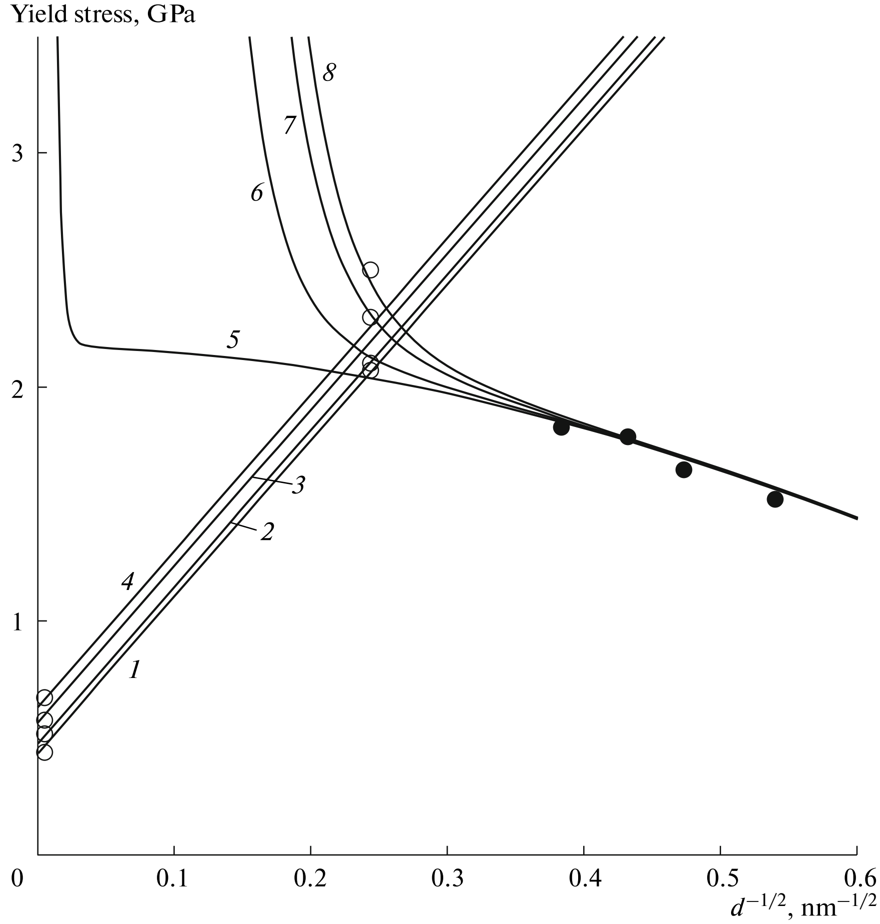

Fig. 2. (*1–4*) Grain-boundary hardening (Hall–Patch dependences) and (*5–8*) softening ("back" Hall–Patch effect) for nickel at four different strain rate: (*1*, *5*) 0.001; (*2*, *6*) 1000; (*3*, *7*) 3000; and (*4*, *8*) 4500 s^{-1}. Symbols correspond to experimental data from (open circles) [4] and (solid circles) [15].

Coble mechanism, the dependence of the relaxation time on the grain size can be obtained using the Maxwell high-viscosity fluid model [13] and the relation for the Coble creep over grain boundaries [14] in the following form:

$$\tau_{gb} = \frac{k_b T d^3}{A_c G b^3 D_{gb} \delta}, \quad (5)$$

where k_b is the Boltzmann constant; T is the temperature; d is the grain size; the parameter A_C is equal to 30–50; δ is the width of the grain-boundary self-diffusion region; and D_{gb} is the coefficient of the grain-boundary self-diffusion of atoms.

The use of the obtained dependences of the characteristic relaxation times on the structural parameters of the material makes it possible to describe the behavior of the yield stress of the material not only when the strain rate is varied, but also when its defect microstructure changes. To describe the grain-boundary slip curve, we use the following relation for the barrier stress [12]:

$$\sigma_y^{gb} = 2\times10^{-2}\frac{G}{1-\nu}\left(1-\frac{\delta}{d}\right)^2,$$

where ν is the Poisson ratio. In the range of large grain sizes, the growth in the yield stress obeys the classic Hall–Patch dependence [11] $\sigma_y^0 = \sigma_0 + k_{HP}d^{-1/2}$, where the stress σ_0 corresponds to the yield stress of a single crystal and k_{HP} is the Hall–Patch coefficient, which is tabulated for the majority of metals. In the general case, the yield stress of a material is determined as the lowest of the two "barrier stresses" that correspond to the dislocation plasticity and grain-boundary slip.

Figure 2 shows an example of the calculation of the Hall–Patch curve for nickel at four different strain rates using the relaxation model of plasticity. The symbols correspond to the experimental data from [4, 15]. The hardening–softening transition in the nanomaterial and the behavior of the curves under dynamic loading can be seen. Thus, the consideration of the parameters of the barrier stresses and the characteristic relaxation times, which correspond to different mechanisms of plasticity, makes it possible to predict the unusual behavior of the material when its microstructure transforms in a wide range of strain rates.

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