

ON THE KINETICS OF DYNAMIC STRAIN AGEING

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The stress-strain phenomena for the regime B of the Portevin-Le Châtelier effect (i.e. hopping bands of fast deformation) are analyzed. The depth of dynamic strain ageing (DSA), defined as the maximum stress increase of negative rate dependence due to DSA, is presented for the tensile deformation courses made at various applied rates. The connection of the depth of DSA to the solute- and strain-hardening mechanisms is derived. It is concluded that both the friction as well as the deformation part of the flow stress are appreciably raised by dynamic strain ageing.

KINETIKA DYNAMICKÉHO STÁRNUTÍ

V práci jsou analyzovány napěťové a deformační jevy v režimu B Portevinova-Le Châtelierova jevu (přeskakující pásy rychlé deformace). Pro tahové zkoušky při různých vnějších rychlostech je určena hloubka dynamického stárnutí (DSA), definovaná jako maximální vzrůst napětí v režimu záporné rychlostní citlivosti, způsobená DSA. Je odvozen vztah závislosti hloubky DSA na mechanismech příměsového a deformačního zpevnění. Dochází se k závěru, že dynamické stárnutí prokazatelně zvyšuje jak příměsovou, tak deformační složku skluzového napětí.

1. Introduction

In certain ranges of temperature and strain rate the stress-strain curves of many solid solution hardened alloys exhibit serrated yielding (jerky flow, plastic instability) associated with the Portevin-Le Châtelier (PLC) effect. The occurrence of serrations on the stress-strain curves is indicative of dynamic strain ageing (DSA) which lowers the strain rate sensitivity S of the flow stress. Thus, S may become negative and this is related to the appearance of the PLC effect [1-4]. A great deal of models explaining the PLC effect are based on the assumption that the serrations are a result of the dynamic interaction between gliding dislocations and mobile (diffusing) solute atoms, i.e. serrations are related to the operation of the dynamic strain ageing mechanisms. Moving dislocations are temporarily arrested at obstacles that may be overcome by thermal activation. During this waiting time t_w , DSA causes an additional pinning of dislocations by rearrangement of solute

atoms; an increase in the slip resistance occurs. Then, an additional activation free enthalpy due to the advent of pinners during the waiting time (i.e. the ageing time) should be taken into account. The time dependence of the activation free enthalpy depends on the actual ageing kinetics [4–7].

A number of models (see e.g. [1, 8]) have been presented according to which the occurrence of the PLC effect at a certain strain is caused by the formation of solute atmospheres on the glide dislocations by bulk diffusion. The concentration of vacancies as well as the density of mobile dislocations are assumed to increase as a power of strain. However, it should be mentioned that the vacancy concentration does not significantly influence the initiation of serrations [2, 9, 10]. In some models [2] the occurrence of serrations is explained without any assumption about that vacancies created during plastic deformation contribute to the diffusion process [2, 10–13].

The flow stress, necessary for the dislocation movement, may be expressed as a sum of two components

$$\sigma = \sigma_d + \sigma_f, \quad (1)$$

where σ_d is the contribution of dislocations-dislocation interactions to the flow stress and σ_f is the friction stress due to the interaction between the solute atoms and moving dislocations. Mulford and Kocks [2] have assumed that it is the dislocation stress σ_d rather than the friction stress σ_f that is influenced by DSA. According to their model the moving dislocation temporarily arrested at the obstacles absorb solute atoms due to pipe diffusion along the forest dislocations. This way, the dynamic strain ageing may correspond to strengthening of the nodes between mobile and forest dislocations. On the other hand, Balík and Lukáč [12, 14, 15] have analyzed the motion of dislocations in the field of solute atoms that also move in the lattice. In their model the dynamic strain ageing phenomena are ascribed to local changes of solute atom concentration, which influence the dislocation velocity and the friction stress.

More recently, Kubin and Estrin [13] have proposed a model explaining the critical strains for the appearance of plastic instability in terms of the strain evolution of the densities of mobile and forest dislocations. In their model the explanation of the occurrence of the PLC effect is not connected directly to a choice of the DSA mechanism.

Very important results concerning dislocation motion during serration yielding have been obtained by a combination of tensile testing with acoustic emission measurement [16–19]. The observed acoustic emission activity is found to be closely connected with the occurrence of serrations. The acoustic emission activity observed is due to the dynamic collective motion of dislocations. Similar conclusions follow from the observations of the specimen surface using optical microscopy and a high-speed camera [20, 21].

In the present work an attempt is made to consider the influence of solute atoms on both the dislocation stress and the friction stress in DSA regime.

2. Basic relations

Using the Orowan equation the strain rate is given by

$$\dot{\epsilon} = b\rho_m\Lambda/t_w = \Omega/t_w, \quad (2)$$

where b is the magnitude of the Burgers vector, ρ_m is the mobile dislocation density, Λ is the mean free path of dislocations. If forest dislocations are the rate controlling obstacles then the elementary plastic strain (per activation event) $\Omega = b\rho_m\rho^{-1/2}$, where $\rho \cong 1/\Lambda^2$ is the density of forest dislocations, is strain dependent. If we consider the thermally activated cutting through forest dislocations as the rate controlling process, the plastic strain rate may be expressed by an Arrhenius law

$$\dot{\epsilon} = \Omega\nu_0 \exp(-G/kT), \quad (3)$$

where ν_0 is the attack frequency, k is the Boltzmann constant, and T is the absolute temperature. The activation free enthalpy G depends not only on the stress acting on the arrested dislocation segment but also on the actual DSA stage.

The activation free enthalpy for overcoming of an aged dislocation segment is given by

$$G = G_0(F) + \Delta G(F, \Delta c), \quad (4)$$

where $G_0(F)$ is the activation free enthalpy corresponding to the thermally activated process in the absence of DSA, F is the actual pinning force. The local concentration increment Δc can be expressed as [5, 22]

$$\Delta c = c - c_0 = \Delta c_M [1 - \exp(-t_w/t_0)^p], \quad (5)$$

where c is the local solute concentration in the dislocation core, c_0 is the nominal solute concentration in the lattice, and Δc_M is the maximum (buffer) concentration increment. The exponent $p = 2/(n + 2)$ and $1/(n + 2)$ for bulk and pipe diffusion, respectively. The value of n depends on the details of interaction between dislocations and solute atom [6, 23]. The relaxation time t_0 depends on the binding energy between a dislocation and a solute atom, on solute concentration, and on the diffusion coefficient of solute atoms. t_0 is inversely proportional to the diffusion coefficient in the case of bulk diffusion whereas for pipe diffusion $1/t_0 \sim D\rho^{(n+2)/2}$ [23].

The force profile $\Phi(x, c)$ (x is the displacement of a dislocation) for cutting mobile dislocations through forest dislocation is assumed to be modified by the local solute concentration c . We suppose $\Delta\Phi = \Phi_0\kappa\Delta c$ where Φ_0 is the force profile corresponding to the nominal concentration c_0 and κ is a constant. Using the Arrhenius law and the relation $\sigma_d = F/bL \approx F\rho^{1/2}/b$ (L is the mean distance between obstacles along the waiting dislocation) the following equation for the dislocation component of the flow stress can be derived

$$\sigma_d = \sigma_{d0}(\dot{\epsilon}/\nu_0\Omega, T, \rho, c_0) + \Psi(\dot{\epsilon}/\nu_0\Omega, T, \Delta c)\sigma_M(\rho), \quad (6)$$

where $\sigma_M(\rho)$ is the low-temperature or high-rate limit of the dislocation stress. This quantity depends on the dislocation density and can be used as a suitable structure parameter, $\sigma_M \approx \sigma - \sigma_0$. If we assume that the shape of the force profile is a triangle, then it holds

$$\Psi = \kappa\Delta c - \beta^{1/2}[(1 + \kappa\Delta c)^{1/2} - 1] \approx 2\kappa\Delta c/3, \quad (7)$$

where $\beta = (2kT/mF_M^2) \ln(\nu_0\Omega/\dot{\epsilon})$, F_M is the mechanical maximum of the force F , and $m = \partial^2 G_0/\partial F^2$ is the change of the activation distance per force unit. The averaging over $0 < \beta < 1$ has been used in the last estimate in (7).

In the case of the friction stress the force profile is not influenced by the concentration of solutes. The Arrhenius equation has the following form

$$\dot{\epsilon} = \nu_0\omega \exp[-g(f)/kT], \quad (8)$$

where the activation free enthalpy g for the dislocation-solute atoms interaction depends on the actual pinning force f , and $\nu_0\omega$ is a relevant pre-exponential factor. The effective solute spacing l along the waiting dislocation segment may decrease due to DSA. It means that the local concentration $c = b/l$ increases. Then, one can derive the following equation for the friction stress

$$\sigma_f = f/bl = \sigma_{f0}(\dot{\epsilon}/\nu_0\omega, T, c_0) + f(\dot{\epsilon}/\nu_0\omega, T)\Delta c/b^2. \quad (9)$$

The actual force f may depend on strain (deformation) only through the factor $\nu_0\omega$, see Eq. (8).

3. Discussion

Using the superposition law (1) and the expressions (6), (9), the decomposition of the flow stress on a non-aged part σ_0 and a DSA-part σ_a is available:

$$\sigma = \sigma_0 + \sigma_a, \quad (10)$$

where

$$\sigma_0 = \sigma_{d0} + \sigma_{f0}, \quad (11)$$

$$\sigma_a = \Psi(\dot{\epsilon}/\nu_0\Omega, T, \Delta c)\sigma_M(\rho) + f(\dot{\epsilon}/\nu_0\omega, T)\Delta c/b^2. \quad (12)$$

For the purposes of estimation, we can use the last expression in (7) for the Ψ -factor. The strain rate dependence of the DSA-stress increment is then simply related to the kinetic law $\Delta c(t_w)$ for the local concentration changes:

$$\sigma_a = [(2/3)\kappa\sigma_M + f/b^2]\Delta c(t_w). \quad (13)$$

Combining the typical kinetic law (5) for Δc with the relation (2), the negative strain rate dependence of the DSA-contribution σ_a is obvious. The form of the DSA-amplitude

$$f_0 = [(2/3)\kappa\sigma_M + f/b^2]\Delta c_M \quad (14)$$

is then similar to that discussed by Van den Beukel and Kocks [24].

The first terms in Eqs. (12–14) result from ageing of dislocation nodes; (this corresponds to the dislocation-dislocation interaction influenced by DSA). They increase with σ_M , i.e. they are deformation depending. The second terms in Eqs. (12–14) result from ageing of dislocations segments between forest dislocations; these terms reflect the solute atoms-dislocation interaction influenced by DSA and they are expected to be practically strain independent.

The strain rate sensitivity of the flow stress is implied by both the sensitivity of non-aged part σ_0 (Eq. (11)) and the one of DSA-contribution σ_a (Eq. (12)). If a region with the negative total strain rate sensitivity appears, the rate instabilities of plastic deformation, called Portevin-Le Châtelier effect, arise.

In order to discuss dynamic strain ageing, we have analyzed the stress-strain phenomena in the regime B of the PLC-effect for an Al-3Mg alloy. The samples were tensile tested at room temperature; plastic dilatation Δl_p of the gauge length l_0 , plastic dilatation Δs_p of an extensometer base δ and force F were simultaneously recorded. Typical motives of such signals are shown in Fig. 1; S_0 is the initial cross section of the sample. The burst-like course of plastic deformation and space sequentiality of plastic events are obvious. The width w of a single plastically

active zone as well as local strain increment $\Delta\varepsilon$ could be evaluated from the above presented records (Fig. 1). For details and complete results see [25].

As the homogeneity of state in the limit of a single plastically active zone is commonly accepted, we omit any gradient terms and rewrite the kinetic law (10–12) in the form introduced by Penning [26] (see also [27])

$$\sigma = h\varepsilon + \tau(\dot{\varepsilon}). \quad (15)$$

Here h denotes the strain hardening coefficient and $\tau(\dot{\varepsilon})$ is an N -shaped function describing the strain rate sensitivity of the stress. In a certain strain rate range, between the maximum τ_{\max} and the minimum τ_{\min} of the function τ , the derivative of the function is negative. The value of $\tau_0 = \tau_{\max} - \tau_{\min}$ may be referred to as the DSA-depth. This quantity can be extracted from our mechanical experiments: the principal relation $\tau_0 = h\Delta\varepsilon - \Delta\sigma$ ($\Delta\sigma$ is the local stress drop accompanying the strain burst) reads in accessible quantities

$$\tau_0 = (h - F_i/S_0)\Delta\varepsilon - (1 + \varepsilon_i + \Delta\varepsilon)\Delta F/S_0,$$

where ΔF is the load drop accompanying the strain burst, F_i is the maximum load at the beginning of the strain burst and ε_i is the corresponding local pre-strain; we take it as the momentary value of $\Delta l_p/l_0$.

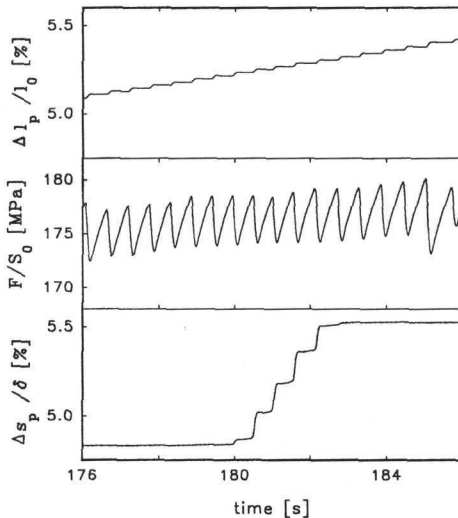


Fig. 1. Plastic phenomena for Al-3Mg samples in the PLC-region. Gauge length $l_0 = 50$ mm, base length $\delta = 5$ mm, initial cross section 6×1.5 mm, applied rate $3.3 \times 10^{-4} \text{ s}^{-1}$.

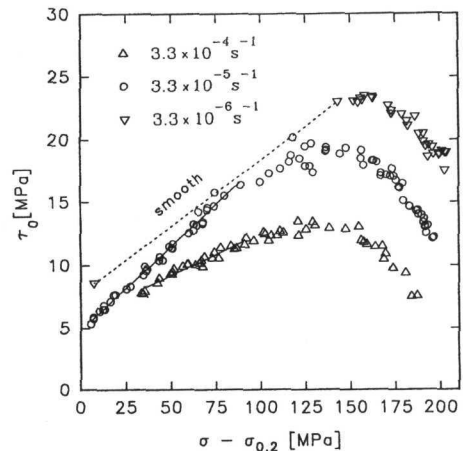


Fig. 2. The depth of dynamic strain ageing for three applied rates.

For the typical kinetics of non-aged flow stress σ_0 and concentration DSA-increase Δc , the DSA-depth τ_0 is nearly proportional to the DSA-amplitude f_0 (14). Thus, the "linearity" of the initial parts of the plots τ_0 versus $\sigma - \sigma_{0.2}$ in Fig. 2 and the obvious positive ordinates of them in the limit $\sigma - \sigma_{0.2} \rightarrow 0$ reveal the dynamic strain ageing influence on both the dislocation contribution as well as on the friction one to the flow stress. The similar conclusion should be valid for the solid solution hardening as a whole.

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