Dislocation Dragging Mechanisms and Long-Term Strength of Metal Crystals

Vladimir G Tkachenko*
I.M. Frantsevich Institute for Problems of Materials Science, Ukraine

Abstract
A new first-order physical theory of long-term strength based on a dislocation model and analytical rate-equations has been advanced for describing dynamic (time-dependent) microyield/creep resistance responsible for a potentially useful measure of the uniform strain preventing a premature fracturing of rapid-hardening crystals. Such a diagnostic approach enables the short-range rate-controlling mechanisms to be identified for stressed crystals in terms of the thermoactivation (numerical) analysis of rapid strengthening using the constant structure steady-state creep tests and dislocation relaxation technique. For the alloys under consideration, a more accurate (dislocation) criterion of useful long-term strength is formulated for solution-hardened alloys in which the threshold dragging stress as a function of uniform strain resistance is directly related to the elastic (shear) stability of a dislocated crystalline lattice, velocity, and density of sliding dislocations as well as their line tension in the stress field. The criterion could be used for the quantitative assessment of the short-range dragging effects preventing a transition from the structurally uniform sliding to the localized shear strain that governs a premature fracture of metal crystals.

Keywords: Strength; Diagnostic approach; Dislocation resistance

Abbreviations: DSA: Dynamic Strain Aging; PLC: Portevin-Le Chatelier; AE: Energy of Activation; AV: Activation Volume

Introduction
Today and even more so in the future, microyield/creep in shear appears to be one of the fundamental material properties and key factors limiting service conditions for a new generation of metal alloys [1-4]. The increasing dynamical application of solution-hardened alloys as cost-effective structural materials stimulates extensive research to reveal the rate-controlling mechanisms responsible for improving their pre-yield (creep) resistance and long-term strength below macroscopic yield stress [4-6]. Nevertheless, over the years, research works under review [7-9] have been mainly performed to study the macroscopic variations in strain for a better understanding of the serrated yielding as a macroscopic phenomenon arising from the dislocation-solute interaction by a long-range diffusion-controlled mechanism. This type of discontinuous and repeated yielding commonly referred to as the Portevin-Le Chatelier (PLC) effect is attributed to dynamic strain aging (DSA), in which the isolated solute atoms interact with mobile dislocations during straining. However, up to now, there is no evidence of a relationship between PLC banding and shear banding [9,10]. According to Cottrell concept [11], the DSA effect has firstly been associated with the dragging stress being caused by the formation of dense solute atmospheres around moving edge dislocations at the same velocity of alloy constituents. DSA in solid solutions is described as the diffusion of solutes to mobile dislocations temporarily arrested at obstacles [8]. It was suggested that a PLC effect expected from solute atmosphere dragging [12] arises from the dynamically unstable region with the jerky flow and serrated yielding subsequent to its formation. Following Cottrell’s concept of jerky flow, the thermally activated process of serrated yielding should occur with the activation energy (a.e.) for solute migration towards dislocations by a diffusion-controlled mechanism. In contrast, the a.e. of the PLC effect which controls the appearance of the serrations is comparable to vacancy migration energy, while the a.e. which governs the disappearance of the phenomenon is similar to self-diffusion energy [7].

There have been several attempts [13-15] to modify the Cottrell concept and to ascertain a DSA effect and jerky flow origin but neither of them fails to explain quantitatively the
rapidity of the hardening in metals with decreasing temperature. Besides, much of on-going theoretical models describe the related PLC and DSA effects at long-term loads close to or above macroscopic yield without respect to the observed short-range interaction of point defects [15-18]. Much as these works have contributed to the elucidation of this problem but our knowledge of it is still far from being complete. At last, there is a long-standing argument that macroscopic changes in strain are accompanied by the multiplication of dislocations so that the thermoactivation (numerical) analysis is certain to become impossible by reason of the fine structural variations.

It has been until very recently that the basic assumptions of the Cottrell theory [11] have been seriously called in question [19,20]. The dislocation atmospheres forming and the bound dislocations dragging are considered to occur during microyield at stresses \(d\tau\/d\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\nu\numb
where \( \tau_0 \) is a frequency factor dependent on lowering a potential barrier; \( U^* \) - the a.e. needed for overcoming a thermally activated obstacle by a short-range mechanism. The relation (3) for an atom in the field of dislocation is different from well-known relation (2) by an amount of the binding energy \( E_b \)

\[
\tau = \tau_0 \exp\left(\frac{U^* + E_b}{k T}\right) \tag{3}
\]

were \( E_b \) is the binding energy of a solute atom, (a paired defect or an associate) with a dislocation.

A little manipulation of equations (1) and (3) yields a dragging effect, if a single activation process controls the microyield/creep resistance and the number of places where the strain per successful fluctuation can take place. The rigorous description of the stress-induced dragging effect in terms of dislocation physics leads to two types of the differential kinetic rate-equations which represent the related phenomena such as dislocation dragging during incipient pre-yielding and pre-yield (dislocation) relaxation.

In the \( \tau^* \cdot V^* > k T \) pre-yield range, i.e. at higher stresses and elevated temperatures, there is no breaking off a dislocation from its atmosphere due to strong binding energy \( E_b \) between them. By reason of enough pinning of dislocations, the height of potential barrier \( U^* \) increases by the value of \( E_b \), i.e. \( U^* \Rightarrow E_b \), with a fraction of the line defects proportional to \( \exp(E_b/kT) \). Then for describing thermally activated dynamical dragging by the dislocation atmospheres, the equation for the shear micro strain rate can be reduced to the following form

\[
\dot{\varepsilon}_s = 2 \rho_s (b \lambda \nu) \exp\left(\frac{-U^* + E_b - \tau^* \cdot V^*}{k T}\right) \tag{4}
\]

where \( \rho_s \) is equated to \( 2 \rho_s (b \lambda \nu); \) \( \rho_s \) is the average mobile dislocation atmospheres density; \( b \) - the Burger′s vector; \( \nu_0 = \lambda \cdot \nu \) - the velocity of dislocation atmosphere dragging; \( V = L \cdot b \) - the activation nano volume; \( L \) - the activation length of mobile dislocation nano segment between mobile nodes of pinning; \( \tau^* \) - the thermal component of applied stress.

In the \( \tau^* \cdot V^* < k T \) pre-yield range, i.e. at lower stresses and higher temperatures, in crystals containing solute atoms of low pinning strength [23], the binding energy \( E_b \) makes it possible the jump-like movement of dislocation to be typical for the onset of jerky flow by strain to first jerk. Therefore, the energy-level of ground state for the potential barrier \( U \) might be split up into two energy-level substates with the increased a.e. of \( U \) and \( E_b \).\( \tau \) is one of the important parameters controlling dislocation creep resistance in the solid solution systems with mobile defect structure.

At higher stresses and elevated temperatures \( (\tau^* \cdot V^* < k T) \) the dislocation creep resistance for a given magnesium alloy in the Mg-Al-Ca-Ti system is found to be essentially governed by a rate-controlling mechanism operating with the a.e. of 0.80±0.02eV or 1.28±0.10eV for excess vacancy migration in the stress field of dislocation atmospheres (Figure 1). In addition, the value is in agreement with those estimated for some alloy crystals [8,23,25,26] and constitutes roughly one-third of that for self-diffusion (2.08±2.24x10^-11). Snoek locking [27] is believed to be the stress-induced ordering mechanism operating in these alloys with a high rapidity due to the migration of excess vacancies to form solutes atmosphere around moving dislocations. Strong pinning of dislocations should be expected when locally bounded matrix atoms find themselves part of a passing dislocation [25]. Aging is as being caused by the movement of the vacancy around immobile solutes and as a sequence by the rotation of solute-excess vacancy pairs to form the segregations around moving dislocations [12]. The rate-equation (4) confirms the rapid nature of short-range interaction between the dislocations and solutes in solid solutions enriched by excess vacancies. According to the analytical estimations of rapid kinetics in the \( \tau^* \cdot V^* < k T \) range the dislocation atmospheres are certain to generate in magnesium alloys the strain-produced excess vacancies.

At lower stresses and higher temperatures \( (\tau^* \cdot V^* < k T) \) the dislocation creep resistance is assumed to be controlled by a dragging mechanism operating with the a.e. of 0.50±0.02eV or 0.85x10^-11 for migration of excess vacancy/solute paired complexes in the stress field around mobile dislocations. Its reliable value yielded by the solution of the Eq. (5) is significantly lower than that for self-diffusion (1.35eV) and lower than that for isolated vacancy formation (0.8eV) and diffusion of any alloying element, i.e. another most probable dislocation mechanism is rate-controlling. As fast as an atmosphere consisting of the complexes

\[
\dot{\varepsilon}_s = 2 \rho_s (b \lambda \nu) \exp\left[\frac{U^* - \tau^* \cdot V^*}{k T} - \sinh\left(\frac{E_b}{k T}\right)\right] \tag{5}
\]

Where, kT is the thermal energy of a solvent lattice.

The original idea for dislocation analysis is that the thermally activated energetic level is split into two sublevels of the short-range potential obstacle due to the discontinuous (repeated, positive and negative) increment of internal stress \( \pm \Delta \tau \). The latter arises out due to the two dragging forces needed for the thermally activated depinning and repeated thermally activated pinning of dislocations [24] to satisfy the Eq. (5) and to the Le Chatelier-Braun principle. From the physical point of view, over the \( \tau^* \cdot V^* < k T \) range splitting of the energetic potential means its growth \( U^* + E_b \) at \( \nu_0 < \nu \), and its decrease \( U^* - E_b \) at \( \nu_0 > \nu \) [12,21]. Thus, the binding energy \( E_b \) is one of the important parameters controlling dislocation creep resistance in the solid solution systems with mobile defect structure.
is formed around a dislocation, the activation short-range ordering mechanism in the stress field of dislocations enables dragging effect to be enhanced much more compared to single solute atoms. Taking into account that the unlocking of pinned dislocations and the motion of unpinned dislocations are controlled by the same thermally activated mechanism [27,28] the thermally activated slip dragging by the excess vacancy/solute complexes having a high rapidity of the short-range pinning is believed to be the most likely rate-controlling dragging mechanism responsible for the pre-yield resistance.

Figure 1a & 1b: Activation volume $V^*$ as a function of applied stress thermal component $\tau^*$ (a) for a dislocated solid solution in hcp Mg-12.5Al-1.3Ca system (1, 2, 3) compared with that of the fcc Fe-32Ni-0.1C system (4): 1-dynamical dragging of mobile dislocations by solute-excess vacancy atmospheres with the a.e. of 0.8 at 423K evaluated by Eq. (4); 2-discontinuous (repeated) pinning of dislocations in early stage of jerky flow conditions with the a.e. of 0.5eV at 523K evaluated by Eq. (5); 3-dynamical dragging of dislocation by solute atmospheres under jerky flow for Fe-32Ni-0.12C alloy [23]; variations of activation volume $V^*$ with the activation length of mobile dislocation nano segment $L_C$ (b). Here $\nu = 10^{-9}$Hz; $b=3.2 \times 10^{-10}$m; $E_b = 0.3$ eV, $kT=0.036$ eV at 423K and 0.045 eV at 523K; $\varepsilon_0 = 10.24$ s$^{-1}$; $L_c = 6.0$nm (for $V^*=20b^3$), $L_c = 10.8$nm (for $V^*=36b^3$).

Numerical analysis of stress (dislocation) relaxation

Stress relaxation is determined by the following relation Ref. [27]

$$\frac{dr^*}{dt} = -K_{\mu} \epsilon^* (\frac{U^* + E_s - r^* \cdot V^*}{kT}) \tag{6}$$

Where, $t$ is the relaxation time for applied stress; $K_{\mu}$ is the elastic rigidity in tension to be constant at any test conditions. For the thermally activated dislocation mechanisms of dragging by solute atmospheres at $r^* \cdot V^* > kT$, the parameters of relaxation can be determined as follows. By combining Eqs. (1), (4), and (6) one can obtain the differential relaxation equation (7) of the form

$$-\frac{dr^*}{dt} = K_{\mu} \epsilon^* \exp \left( -\frac{U^* + E_s - r^* \cdot V^*}{kT} \right) \tag{7}$$

We omit here some intermediate computations involving the mathematical treatment of Eq. (7) by the combined operations of separating the variables, and setting up the relevant integral equation to meet conditions $r(t)=r_{\max}$ at $t=0$ and $r(t)=r_c$ (final time of relaxation). Then this relaxation equation was integrated and found the logarithm. After a little manipulation, the expression obtained was re-arranged as the equation of dislocation relaxation which can be described by the expression

$$\Delta r^* \cdot V^* = U^* + E_s - kT \ln(\beta \cdot t) \tag{8}$$

where $\beta = \frac{K_{\mu} \varepsilon^*}{kT}; \varepsilon^* = 2 \rho^* b \lambda V^*; V^* = L_c b^3; L_c = 2r/\varepsilon_0 \cdot \exp(-E_b/kT)$; $L_c$ is the length of dislocation nanosegment between mobile nodes of pinning; $r$-radius of curvature for a dislocation $(r=Gb/\tau); \varepsilon_0$-initial concentration of solutes.

Taking the antilogarithm of Eq. (8) we get the modified equation for the relaxation time

$$t = \frac{1}{\beta} \exp \left( \frac{U^* + E_s - \Delta r^* \cdot V^*}{kT} \right) \tag{9}$$

In going from thermal to a thermal component $\tau_{se}$ (at $\tau^*=0$) the Eq. (8) of stress relaxation reduces to the second-kind equation of relaxation

$$U^* + E_s = kT \cdot \ln(\beta \cdot t_{se}) \tag{10}$$

By analogy with the above-mentioned procedure, the stress relaxation equations can also be deduced
for the discontinuous (repeated) pinning of dislocations in the \( \tau^*V^*', kT \) range. For the case, the course of calculations is almost the same as that obtained for the \( \tau^*V^'>kT \). Then the dynamic dragging of mobile dislocations is described by the relation

\[
\Delta \tau^*V^* = U^* - kT \ln(\beta) - C \ln \left( \frac{\dot{\varepsilon}}{kT} \right)
\]

(11)

The present stress relaxation analysis is found to fit generally into semi-logarithmic relationships obtained previously by [29,30]. Nevertheless, numerical analysis of the equations in the early models of relaxation [29-31] is absent which makes it difficult to examine their reliability and justification for their analytical solutions. At the same time all the terms in our newly developed equations of stress (dislocation) relaxation have physical meanings.

A pronounced solid solution strengthening effect is attributed to the large long-range stress fields due to a thermal stress \( \tau_c \) while the relative strength of the short-range obstacles produced by solutes is not so obvious. The energy of dislocation, when unaided, is insufficient to pass a tetragonal (paired) defect as a short-range barrier. The extra energy to overcome it must be supplied by thermal fluctuations. The present theory enables the net a.e. needed for overcoming the height of short-range barrier potential to be resolved into two time-dependent constituents associated with the partial a.e. for power (elastic) interaction between alloy ingredients, and the partial a.e. for long-term thermal vibrations of a lattice (Figure 2). The observed logarithmic plot of stress (dislocation) relaxation at each temperature is certain to consist of the two linear segments separating the data into thermal and power (elastic) energetic contributions. The two energy-sharing components acting at atomic distances (-0.1nm) may be exemplified by two energy terms: (i) the first term \( \Delta \tau^*V^* \) is due to the work done by the drag stress component \( \tau' \) and (ii) the second term \( U^* - \Delta \tau^*V^* \) in Eqs. (4) and (5) should be considered as energy of thermal vibrations for a crystal lattice. Thermal energy as a driving force equal to the shaded area on the diagram will move dislocation on a short-range distance by the energy of atomic thermal fluctuations linked to the crystal lattice vibrations. The idea of separating the time-dependent contributions concerning the thermal and mechanical breaking of dislocation is consistent with the concept of thermally activated dislocation depinning by [24]. Numerical analysis of the dislocation relaxation following Eqs. (8) and (11) shows that the power component \( \tau' \) decreases linearly with loading time and tries to be as high as the a thermal component \( \tau_c \) (at \( \tau'=0 \)), while the energy when supplied to the short-range obstacle by long-term thermal fluctuations, on the contrary, increases linearly with time up to the value of resistance to the dislocation motion. The ordering forces exerted by a stress field around dislocation are opposed by the disordering forces of lattice thermal vibrations [23], the thermal energy of which tends to randomize a solute distribution. The growing contribution of thermal activation energy (Figure 2) signifies the increase in the time-dependent strain produced by successful thermal fluctuations. This theoretical calculation is a good test for the physical sense of the model proposed.

Energy (dislocation) criterion of useful long-term strength

Principles of effective alloying have been quite successful in the past in developing alloys for high-temperature service aerospace applications. Following concept [1] a measure of “maximum usefulness temperature” during straining is arbitrarily defined as the temperature at which the alloy can withstand the stress of about 69MPa for 100 hours without fracturing. This qualitative criterion allows evaluating the alloy-efficiency for pure metals. Over the past several years, in applied physics successfully advanced in terms of preceding dislocation theory, on the whole, a new theoretical-cognitive situation has shaped which requires the

![Figure 2: Numerical analysis of the dislocation relaxation for HCP Mg-12.5Al-1.3Ca0.1Ti alloy with separating the time-dependent contributions of the power \( \Delta \tau^*V^* \), (1 and 2) and thermal \( U^* - \Delta \tau^*V^* \), (1′ and 2′) components to the net activation energy for overcoming a short-range potential obstacle by a dislocation at 523K (1 and 1′) using Eq. (11) and 423K (2 and 2′) using Eq. (8) in nano volumes 36 b and 20 b, respectively. Here \( U \) is the full free (available) energy for thermally activated overcoming the short-range obstacle; \( K_e=4 \times 10^6\ J/m^2 \); \( G=17\ GPa \); \( \mu=0.35 \).](image)
development of more accurate (quantitative) criteria for further improving physico-mechanical properties of metal crystals below the macroscopic yield stress [3,10,19,32]. Following the situation, it is much appropriate to develop a diagnostic approach that provides for the opportunity of predicting and preventing the harmful and dangerous defects in solution-hardened metal alloys for their safe long-term exploitation in machinery. A little manipulation yields energy balance for the steady-state of dislocation creep 

\[ \varepsilon_i = k \cdot i \] and for the characteristic time \( t' \) to achieve a measured quantity of useful (uniform) strain before a localized shear, i.e. ( \[ \Delta \varepsilon_{\text{r}} : \varepsilon_{\text{r}} = E_0 \cdot \rho_0 \cdot \Delta \varepsilon_{\text{r}} \]), and \( t' = E_0 / \Delta \varepsilon_{\text{r}} \cdot \rho_0 \cdot \Delta \varepsilon_{\text{r}} \). This makes it possible to combine the equation for the linear energy of dislocation \( E_0 = \frac{G b^2}{2} \frac{[\{4 \pi (1-\mu)]}{8 \pi (1-\mu)} \cdot \rho_0 \cdot \Delta \varepsilon_{\text{r}} \) and Eq. (1) for \( \dot{\varepsilon}_0 \) with Eqs. (4) for strain rate, \( \dot{\varepsilon}_m \) at \( \Delta \varepsilon_{\text{r}} : \varepsilon_{\text{r}} > kT \). For the case, the general ultimate transition to the elastic instability should be replaced with a newly developed dislocation criterion of instability for a defect structure as

\[
\Delta \varepsilon_{\text{r}} : \varepsilon_{\text{r}} \sim \frac{g b \varepsilon}{8 \pi (1-\mu) \rho_0} \exp \left( \frac{U + E_0 - \varepsilon_{\text{r}}}{kT} \right) \quad (12)
\]

\[
\Delta \varepsilon_{\text{r}} : \varepsilon_{\text{r}} \sim \frac{g b \varepsilon}{8 \pi (1-\mu) \rho_0} \exp \left( \frac{U + E_0 - \varepsilon_{\text{r}}}{kT} \right) \quad (13)
\]

With this provision, at \( T \to T_{\text{us}} \), and \( \varepsilon_{\text{r}} \to 0 \)

\[
\Delta \varepsilon_{\text{r}} : \varepsilon_{\text{r}} \sim \frac{g b \varepsilon}{8 \pi (1-\mu) \rho_0} \exp \left( \frac{U + E_0}{kT} \right) \quad (14)
\]

Here \( \Delta \varepsilon_{\text{r}} \) is the threshold dragging stress, under which a transition from uniform strain to localized shear is the most likely to occur at \( \varepsilon_{\text{r}} \); \( g \)-Schmidt factor; \( \mu \)-Poisson’s ratio; \( Gb \)-the elastic shear stability (rigidity) of crystalline lattice; \( g b \)-the threshold dislocation energy corresponding to the measured shear rate; \( \rho_0 \)-the density of dislocations; \( \varepsilon_{\text{r}} \)-the threshold strain rate; \( T \)-the useful temperature by \[1\]; \( T_{\text{us}} \)-the entropic temperature, under which thermal fluctuation are capable to overcome a short-range potential obstacle without stress \( (\tau = 0) \). According to the criterion, the threshold stress of dragging is related directly with the elastic shear stability of a crystal lattice, strain rate, density, excess energy and velocity of dislocations as well as strength of interatomic bonds (through \( E / G \) modulus). Interconnection of the subcritical parameters \( (\varepsilon_{\text{r}}, \rho_0, \rho_0) \) with various rigidity, i.e. the shear stability of crystal lattice (\( G b \)) and linear tension (excess energy) of dislocations (\( g b \)) enables to predict the attainment of the dislocation strength at short-range distances. The analytical equations \( (12-14) \) with the same structure furnish a consistent criterion of the useful long-term strength connecting the threshold stress \( \Delta \varepsilon_{\text{r}} \) with properties of the crystalline lattice, elastic fields of mobile dislocations as well as with the activation parameters that reveal the nature of a most probable rate-controlling mechanism. Besides, the energy (dislocation, quantitative) criterion describes a subcritical transition from the uniform elongation through a uniformly dense dislocation distribution to the concentrated slipping and localized shear responsible for the loss of the shear elastic stability for crystal lattice and, as a sequence, microplastic (dislocation) instability.

At low strain rates dragging in steady-state increases almost linearly with the dislocation velocity \( \dot{\varepsilon}_0 \) only in the early stage of the solute atmosphere formation. According to the inequality \( (12) \) at \( d \dot{\varepsilon}_0 / d \dot{\varepsilon}_0 < 0 \) the dragging stress is expected to increase with decreasing the dislocation velocity \( \dot{\varepsilon}_0 \), when it exceeds the critical value \( \dot{\varepsilon}_0 \) by a Cottrell solute diffusion-controlled (long-term) mechanism \[12,33\]. Delay time is inversely proportional to the dislocation velocity \( (-1/\dot{\varepsilon}_0) \) and has a maximum at a critical dislocation velocity \( \dot{\varepsilon}_0 \). Hence for strain rate \( \dot{\varepsilon}_0 \) according to Eqs. (12) and (13). In this case, after passing through a maximum at \( \dot{\varepsilon}_0 \) and \( \rho_0 = \text{const} \) due to a partial unlocking of rapid dislocations, the system in the stable state turns into dynamically unstable plasticity state in the region with rapid dislocation sliding. The theoretical results obtained for dislocation velocities by inequality \( (12) \) fitting a measured strain rate \( \dot{\varepsilon}_0 \) by the inequality \( (13) \) are in a good agreement with modeling study and theoretical calculation of critical dragging maximum observed \[12\]. The validity of the present dislocation criterion was experimentally verified using a strain-rate change test which has been described in detail elsewhere \[24,34\]. The creep rates \( \dot{\varepsilon}_0 \) calculated with Eqs. (4) and (5) and compared with experimental data are summarized in Table. It is noteworthy that the contribution of the short-range thermally activated mechanisms to the dislocation microyield/creep resistance is thought to be enhanced with slowing strain rate (dislocation velocity). The newly developed magnesium alloy in the hcp Mg-Al-Ca system is found to have the best combination of short-term and long-term properties compared with commercial alloy AZ91D.

Discussion

A possibility for solute atoms to interact with a dislocation if the time is too short to activate the long-range diffusion-controlled mechanism was first suggested \[33\] and analyzed \[34\]. The most striking advance in the understanding of the rapid interaction of individual solute atoms with individual mobile dislocations was achieved \[35\]. Nevertheless, despite considerable research efforts in the past, in any event, current theories fail to explain with the present evidence the observed rapid kinetics of hardening in solid solutions as well as the deviation of the fundamental equation for short-range order strengthening from experimental data \[36\]. Some good work is underway along these lines, but original diffusion rate-equations can lead to serious discrepancies since the kinetic parameters are no longer consistent with the values taken from the diffusion data. The correctness of such an interpretation is based on several grounds. Firstly, a dragging effect is related to the direct solute-dislocation interaction with binding energy \( E_g \) rather than to the mobility \( D / kT \) by the Einstein relation, i.e. diffusion drift velocity \[17\]. Unlike the theoretical aspects of macroscopic changes of diffusivity \( D \) describing in the form of diffusion drift the macroscopic flows of substance, the binding energy \( E_g \) gives a result of short-range interaction at atomic distances. Secondly, since the diffusion is not necessary constituent or integral part of high-
temperature creep [37], it is difficult yet to calculate the diffusion process numerically due to the vacancy mechanism which needs many unknown quantities. Only a few theoretical approaches to this problem have been advanced in the literature. To illustrate, according to Sherby with co-workers [18], the atom mobility and elastic modulus compensated stress $\sigma / E$ are believed to be two principal factors influencing the creep strain rate $\dot{\varepsilon}^c$ at a given stress. Nevertheless, although at the same diffusion mobility of solutes the excellent correlation between and $\sigma / E$ has been observed in many metal alloy systems, in the suggested rate-equations [18] relevant to the problem a term $f(\sigma / E)$ is yet unspecified function.

Unlike the Cottrell atmosphere, the ordered Snoek atmosphere is formed in the effective stress field of dislocations by a short-range ordering mechanism [28,36], which causes the dragging force for dislocations moving against the energy gradient depending upon their velocity. Since Snoek short-range ordering precedes the Cottrell mechanism of the solute atmosphere formation, the subcritical value of dislocation velocity $V_D$ is attained for high-purity $\alpha$-Fe crystals at $\dot{\varepsilon}^c$ of $10^{-9}$ s$^{-1}$ even at r.t. [28]. This result is in reasonable agreement with the calculated values of $V_D$ for high-purity magnesium alloy crystals under examination at the same dislocation density (Table 1). The Snoek-type dislocation atmosphere is believed to be formed as the complex consisting of an excess vacancy and a solute alloying element. The idea for a local loss of the crystalline lattice stability near-linear defects was first advanced [37] for vacancy clusters and complexes of points defects having elastic distortions around them. With that, the achievement of critical parameters for the store internal energy, e.g., the local critical density of dislocations during time-dependent strain is thought to be caused by deformation instability, first of all, by reason of loss of shear stability of a dislocated crystalline lattice and localized shear below the macroscopic yield stress. These effects are accompanied by decreasing microyield/creep resistance and long-term strength.

### Table 1: Micromechanical properties of conventional and newly developed experimental magnesium alloys at 423K

| Alloy Systems       | Critical Creep Parameters | Critical Creep Parameters | Critical Creep Parameters | Critical Creep Parameters | Critical Creep Parameters |
|---------------------|---------------------------|---------------------------|---------------------------|---------------------------|---------------------------|
|                     | $\sigma_c$, MPa at r.t.   | $\sigma_c$, MPa at r.t.   | $\sigma_c$, MPa at r.t.   | $\sigma_c$, MPa at r.t.   | $\sigma_c$, MPa at r.t.   |
| Mg-9Al-1Zn          | 170                       | 35                        | 200                       | 1.93-2.50                 | $\geq 10^{-7}$            |
| AZ91D (Dow Chemical Corp., USA) | 170                       | 35                        | 200                       | 1.93-2.50                 | $\geq 10^{-7}$            |
| AZ91D-1.3Ca         | 220                       | 64                        | 50                        | 0.1                       | $10^{-7}$                 |
| Mg-12.5Al-1.3Ca     | 157                       | 65                        | 160-180                   | 0.17-0.24                 | $\geq 10^{-8}$            |
| Mg-12.5Al-1.3Ca-0.1Ti | 180                       | 70                        | 150-200                   | 0.17-0.24                 | $\geq 10^{-8}$            |

**Table abbreviations:** r.t: Room Temperature; h: Hours; Exp: Experiment

The appearance of the relaxation spectra of internal friction in hcp magnesium alloy indicates that solute atoms are expected to be paired with excess (strain-produced) vacancies, which attract them strongly producing large (tetragonal) lattice distortions [2,6,33]. With this provision, the vacancies will remain in the solution strongly producing large (tetragonal) lattice distortions [2,6,33].

According to the kinetics of DSA [23] the observed rapid jerky flow on the serrated stress-strain curves can be explained on the basis of the stress-induced short-range ordering mechanism [27,28] rather than Cottrell elastic approximation or Suzuki chemical dislocation pinning mechanism. The rate-equation (5) in subsection 3.1 for repeated pinning of dislocations evaluates the extent of smoothening curves of plastic strain at the initial stage when occurring the first plastic deformation jerks in preceding jerky flow. As a matter of fact, jump-like (discontinuous, repeated) pinning of moving dislocations is likely to facilitate rather than to cause the proceeding of jerky flow or DSA. By reason of its general sense, the magnitude of initial jerks in the form of stress drag increment and abrupt fall in load can be regarded as indicative of the stress differential between the breakdown stress at which dislocations escape from their atmospheres and the stress at which they move easily through the lattice before being pinned again [16]. It is reasonable to assume that in the $T^\ast$ range the dragging effects under examination are typical for the onset of jerky flow by strain to a first deformation jerk the jerkiness effect relates to mesoscopic phenomena to occur below the macroscopic yield stress, our system of the newly developed rate-equations can be expected to give a good result for describing the onset of jerky flow.

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Conclusion

The basic conclusions could be drawn from the results obtained and may be stated as follows:

1. A physical theory of useful long-term strength based on mathematical modeling of the thermally activated (short-range) resistance to dislocation movement and on the Le Chatelier-Braun’s physical-chemical principle of shifting the mobile equilibrium has been evolved for adequate describing solid solutions enriched by excess vacancies and their rapidly-hardened alloys with hcp and fcc structure. The theory furnishes an explanation of the above dragging effects in the early stage of their formation including the onset of atom ordering and describes the direct interaction between defects and solutes involving the jump-like movement of dislocations with increasing temperature. This makes it possible to separate the time-dependent contributions of power and thermal components to the net activation energy needed for overcoming the short-range obstacles.

2. Judging the results obtained, at least two underlying types of the dislocation dragging mechanisms were identified by using a new model-based analytical rate-equations when evaluating the activation energies in intrinsic to creep-resistant alloys in the hcp Mg-Al-Ca, Ti and fcc Fe-Ni-C systems. At higher stresses and elevated temperatures (\(\tau^{\dagger}V^{\dagger}>kT\)) the movement of glide dislocations is most likely to occur jointly with their solute atmospheres so that for the case, the dynamic dragging of micro yielding is provided by the resistance of mobile dislocation atmospheres with the a.e. for vacancy migration. At lower stresses and higher temperatures (\(\tau^{\dagger}V^{\dagger}<kT\)) the dynamical dragging of creep strain is believed to be due to the constant (discontinuous, repeated) interactions between mobile free dislocations and solutes by a jump-like mechanism. For the case, a point peculiar interest is an enhanced effect of dragging in the early stage of the jerky flow. It may be appropriate for improving the long-term strength of high-temperature alloys.

3. Over the time-dependent microyield/creep range, theoretical simulations and experimental evidence predict a potentially useful measure of long-term strength formed in the stress fields of mobile dislocations. The quantitative (dislocation) criterion of useful long-term strength is proposed for quantitative assessment of the alloying effectiveness as to the best possible metal alloy crystals with minimum strain location. The criterion describes material conditions that prevent a transition from uniform strain to localized shear leading to the elastic and microplastic instability of a dislocated crystal lattice. The present theory suggests that the transition state might be considered as being the physical restriction in dislocation long-term strength. In the context, a threshold dragging stress as a function of uniform strain resistance is directly related to elastic (shear) stability of a dislocated crystalline lattice, line tension (excess energy), velocity and density of sliding dislocations.

4. The findings obtained are in reasonable agreement with strain rate change and stress relaxation measurements, test findings and isothermal observations, which are suggestive of the duster-like complexes. In summary, therefore, the present theoretical study will hopefully stimulate further experimental investigations that could lend a better understanding of the mechanisms responsible for the formation of dislocation creep resistance and long-term strength in metal crystals. The present newly developed diagnostic approach is believed to provide a physical basis not only for a proper understanding of dragging effects but also for alloy development. The proposed method can be expected to be appropriate in describing the creep behavior of metal alloy systems as a starting point for further development of the rapid-hardening and creep-resistant alloys with a minimum of localized shear effects.

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