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BACK STRESS IN DISLOCATION CREEP
PART 1. BASIC CONCEPTS AND MEASURING TECHNIQUES

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16. Abstract

A theory is proposed whereby the plastic deformation of metal materials is determined by the difference between the applied stress and the back stress which characterizes the resistance of the material to plastic deformation. The back stress is usually equivalent to internal stress or the friction stress and depends on the magnitude of the applied stress and temperature. The concept of back stress is applied to the case of dislocation creep of precipitation-hardened or disperson-strengthened metal materials. An additivity rule is formulated that can be useful in interpreting the creep behavior of such materials.

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BACK STRESS IN DISLOCATION CREEP
PART 1. BASIC CONCEPTS AND MEASURING TECHNIQUES

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The article advances the idea that plastic deformation of metallic materials at high temperatures is not caused by the entire applied stress σ but only by its part represented by the difference between $\sigma - \sigma_B$, where σ_B is the back stress characterizing the resistance of the material to plastic deformation. The back stress is frequently identified with internal stress σ_i or the "friction" stress σ_0. Both the internal and friction stresses generally depend on applied stress σ and temperature. The internal as well as friction stress can be determined experimentally. In addition, the former can be deduced from model concepts. Techniques of measurement and model concepts are briefly described.

If the back stress σ_B does not depend on applied stress it can be identified with the threshold stress σ_T below which the dislocation creep does not take place. The temperature dependence of threshold stress follows exclusively from that of the elastic shear modulus.

1. Introduction /518*

Considerable acceptance was gained in the course of the past two decades by the concept that deformation of crystalline substances at high temperatures is not caused by the effects of the entire applied stress σ, but merely the effects of its part represented by the difference between $\sigma - \sigma_B$, where σ_B is the back stress. "Pure" stress $\sigma - \sigma_B$, often referred to as effective stress, characterizes resistance of the substance to plastic deformation. The

* Numbers in the margin indicate pagination in the foreign text.

quantity σ_B can depend on temperature, deformation history and, particularly, on applied stress and/or the velocity of deformation. However, not always can it be interpreted as stress encountered by dislocations in creeping motion, even though the adopted adjective "reverse" could imply such interpretation.

The concept that creep velocity can be described by the equation

$$\dot{\varepsilon}_s = A(\sigma - \sigma_B)^n, \quad (1)$$

where A is the function of temperature and n is a constant, formed the basis for many experiments designed to explain the observed dependence of creep velocity on the applied stress and temperature. However, the outlined approach hardly ever led to gaining a deeper insight into deformation processes, undoubtedly because back stress σ_B was not adequately defined for the given conditions of plastic deformation or for the process controlling deformation velocity.

Prior to discussing the possibilities for application of the concept of back stress, we must specify in closer detail the terms internal stress, threshold stress and friction stress, all of which can be included under the general term back stress.

2. Internal, Threshold and "Friction" Stress

Dislocation, propagating by shear in the crystal, encounters local obstacles which it overcomes by thermal activation. For deriving an equation expressing the velocity with which dislocation overcomes local obstacles of the contemplated type [1, 2] it is /519 expedient to adopt the concept of a thermodynamic subsystem with interaction occurring between it and its surroundings. This subsystem includes the obstacle and, with the aid of thermal activation, the dislocation sector undergoes transition from one stable configuration Z into the adjacent stable configuration Z', whereby it passes through configuration U characterized by the maximum of potential energy. In a real crystal this subsystem is subject to the effects of two stresses: the macroscopic homogeneous external (applied) stress σ and internal stress σ_i' caused by dislocations outside of the subsystem at hand. In order for the effect of this internal stress to be taken into consideration, it must be assumed that it is possible to define the homogeneous effective stress in the subsystem along the path from stable configuration Z to stable configuration Z'. Thus, external stress can be divided into two components. The first component which compensates for the effect of dislocations and other eventual sources of internal stress outside the subsystem at hand we shall refer to as internal stress $\sigma_i = -\sigma_i'$. The second component is then constituted by effective stress σ^* that affects the dislocation sector in the selected subsystem becoming active during motion of this dislocation sector. It then applies that

$$\sigma = \sigma_i + \sigma^*. \quad (2)$$

Controlled motion of the dislocation sector is possible only when $\sigma > \sigma_i$. Of course, the dislocation sector can overcome the obstacle even when $\sigma = \sigma_i$, namely by means of thermal activation. However, with probability approaching the order of 1, in the subsequent moment it returns to its original position. At $\sigma < \sigma_i$ the sector is unable to overcome the obstacle, on the contrary, thanks to the effects of internal stress it can move in the opposite direction, i.e., against the effects of external stress.

Internal stress is a complex function of position in the crystal and, of course, the same applies to effective stress as well. In applying the concept of effective and internal stress it is imperative to either: a) estimate in a suitable manner the maximum amplitude of internal stress σ_{i0} and/or minimum effective stress σ^*_{min}, as--in statistical representation--it can be expected that most dislocations in motion at any given moment are situated in areas of low effective stress. Or: b) define the median effective stress. The second of these two alternatives is without a doubt simpler. Let us therefore define the median effective stress, for which we shall retain the designation σ^*, as the average of effective stresses that affect individual dislocation sectors. Then the difference between $\sigma - \sigma^*$ can be designated as "apparent" internal stress (for which we shall retain the designation σ_i and which we will keep referring to simply as internal stress). This internal stress is a measure of the effect of structure--averaged for the entire sample--on the shearing motion of dislocations [3]. Otherwise this (apparent) internal stress can be interpreted as median internal stress encountered by the dislocation during its shearing motion and the source of which are all other present dislocations. Thus, internal stress σ_i can be contemplated only in connection with the shearing motion of dislocations. Internal stress depends on the applied stress and, usually, also on temperature; unless a stationary state is involved it also depends on deformation, of course. From the existence of internal stress σ_i it follows that if applied stress σ is decreased in the course of creep by /520 $\Delta\sigma > (\sigma - \sigma_i) = \sigma^*$, then (some) dislocations will move due to the effects of internal stress against the direction of the effects of applied stress, whch will become manifested by reverse deformation. We shall return to dealing with internal stress σ_i and effective stress $\sigma^* = \sigma - \sigma_i$ later.

If the back stress σ_B does not depend on applied stress, it can be interpreted as threshold stress which we shall keep on referring to by the symbol σ_τ. During applied stresses that are lower than the threshold stress σ_τ creep cannot take place. If in equation (1) $n \gtrsim 1$, then for $\sigma \to \sigma_\tau$ the parameter of stress sensitivity of the velocity of stationary creep $m' = (\partial \ln \dot{\varepsilon}_s / \partial \ln \sigma)_\tau \to \infty$, while for $\sigma \gg \sigma_\tau$ the parameter is $m' \to n$. Thus, the dependence of $\lg \dot{\varepsilon}_s$ on $\lg \sigma$ will be curvilinear (Figure 1).

Wilshire et al. [4-9] introduced the term "friction stress" /521 σ_0 which during creep effectively reduces applied stress so that what has to be taken into consideration is not applied stress σ but the difference between $\sigma - \sigma_0$. The existence of friction stress σ_0

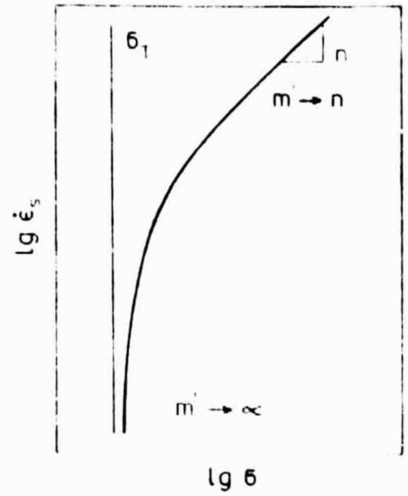

Figure 1. Existence of threshold stress σ_T and dependence on stationary creep velocity on stress in bilogarithmic coordinates. /520

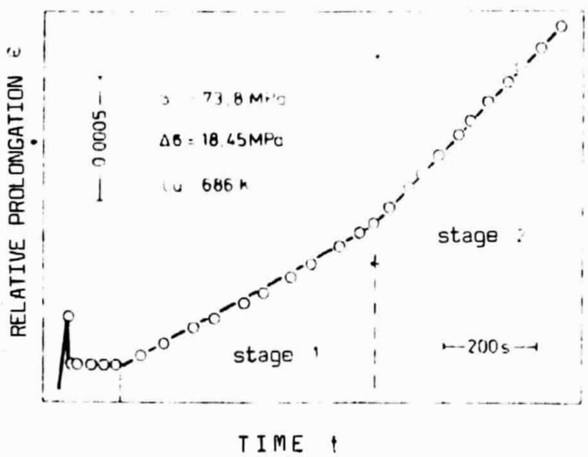

Figure 2. Incubation period following reduction of stress σ; recovery after the incubation period occurs in two stages [4].

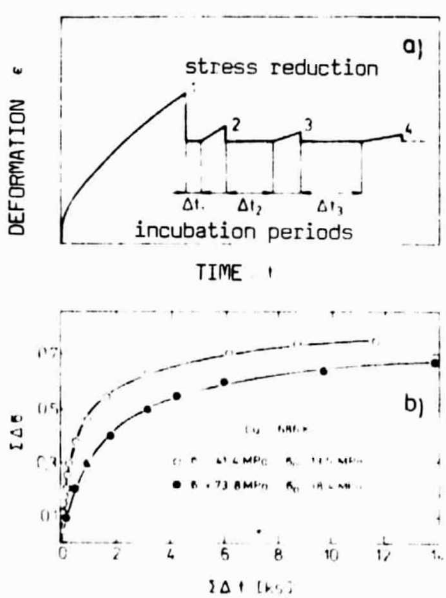

Figure 3. Measuring of friction stress by successive reductions of applied stress. /521

a) Relation between deformation and the time of successive reductions of stress.

b) Relation between cumulated incubation periods and cumulated reductions of stress (copper 686 K (b) [4]).

is tied to existence of the incubation period following a small redution in applied pressure (Figure 2). The technique for measuring stress σ_0 is shown in Figure 3 [4]; immediately after completion of the incubation period lasting for Δt_1 after reduction of stress σ by $\Delta\sigma_1$ is carried out another reduction of stress by $\Delta\sigma_2$ which is followed by an incubation period lasting for Δt_2 etc. (Figure 3a). If then the cumulated reduction of stress $\Sigma\Delta\sigma$ is plotted in relation to the cumulated incubation period $\Sigma\Delta t$, we obtain stress σ_0 as the value of residual stress $\sigma - \Sigma\Delta\sigma$ asymptotically approached by curve $\Sigma\Delta\sigma$ vs $\Sigma\Delta t$, Figure 3b [4]. The technique of measuring stress σ_0 was somewhat modified by Gibeling and Nix [10].

Thus, friction stress σ_0 is residual stress $\sigma - \sum_k \Delta\sigma_k$, during which the cumulative incubation period $\Sigma\Delta t_k$ is theoretically infinitely long.[1] Williams et al. [4-9] interpreted friction stress σ_0 as the characteristics of the dislocation substructure, whereby this dislocation substructure includes the boundaries of subgrains and the tri-dimensional dislocation network within subgrains. Pure stress $\sigma - \sigma_0$ determines creep velocity by determining the velocity of recovery consisting in growth of dimensions of the tri-dimensional dislocation network in subgrains. Thus, the stress σ_0 constitutes the characteristic of the substructure in the sense that it depends on subgrain size and on the overall dislocation density. However, this stress does not depend on the lengths of individual (local) elements of the tri-dimensional network [5], and, thus, is a measure of back stress with a long-range connected /522 with the dislocation substructure. As this stress affects the dislocation network, Gibeling and Nix [11] proposed for it the designation "network back stress" instead of the designation friction stress.

From the presented interpretation it follows that the overall internal stress that affects individual dislocation elements is equal to the applied stress, because these elements do not move until they are released by recovery. As soon as any one of these

[1] $\Delta\sigma_k$ is the k-th stress reduction followed by incubation period Δt_k.

elements is released, it moves very rapidly toward the next obstacle. In this sense the deformation is athermic.

Wilshire et al. [4-9] achieved for value of the exponent $n = 4$ and $\sigma_B = \sigma_0$ in equation (1), even for the case of alloys reinforced by precipitation, an acceptable agreement between the value of the activation energy of creep and the value of the activation enthalpy of lattice diffusion and a satisfactory explanation for the strong (in comparison with pure metals) dependence of creep velocity on applied stress. However, it ought to be once again reiterated that the concept of friction stress σ_0 critically depends on whether the incubation period can be observed (and measured) after small reductions of applied stress. In this respect the experimental results obtained to date are considerably contradictory.

3. Internal and Effective Stress

In the preceding section we outlined the concept of internal and effective stress. In accordance with this concept, applied stress is formed by two components: effective stress σ^* which actually affects the considered dislocation sector and is active during its shearing motion, and internal stress σ_i which encounters this sector in motion. In keeping with equation (2), to the median effective stress corresponds the "apparent" internal stress which we characterized as a measure of the effect of structure, averaged for the entire sample, on the shearing motion of dislocations.

In the case of "well-annealed" single crystals, in which dislocation density is very low, at small deformations ($\varepsilon \leq 0.01$) the difference between effective and applied stress can be disregarded. However, during deformations usually encountered in creep (particularly in its stationary stage), be it in case of single crystals or polycrystals, the density of dislocations tends to be so high that the difference between applied and effective stress cannot be disregarded. Effective stress must then be either measured, or deduced from model concepts.

3.1 Measurement Techniques

If we return to the already mentioned statement, namely that in the selected thermodynamic subsystem the dislocation sector cannot move as long as the applied stress σ does not differ from internal stress σ_i[1], we arrive directly at the principle of two tech- /523
niques for measuring the median effective stress σ^*. These techniques consist in determining the reduction of applied stress σ by $\Delta\sigma = \Delta\sigma_c$ to which corresponds either zero velocity of stress relaxation (stress transient dip test technique), or zero velocity of creep (strain transient dip test technique) measured immediately after reduction of applied pressure. The first of these techniques was proposed independently by Vlach [12] and Gibbs [13] in 1966, the second then by Ahlquist and Nix [14] in 1969. Both techniques are illustrated in Figure 4. The latter shows that if $\Delta\sigma$ is selected so as to correspond to the effective stress σ^*, immediately after reduction of stress the velocity of relaxation $\dot{\sigma}_t$ (Figure 4a) or the velocity of creep $\dot{\varepsilon}_t$ (Figure 4b) is zero. Otherwise the velocity of relaxation of stress or of creep is either positive or negative, depending on whether $\Delta\sigma$ is smaller or larger than the median effective stress σ^*.

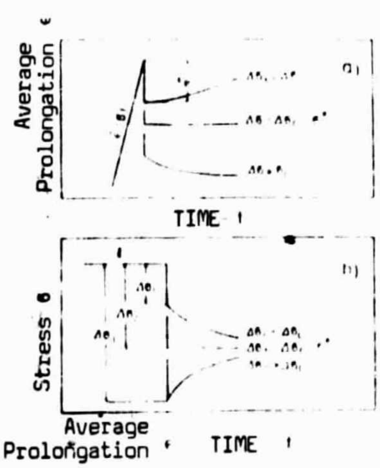

Figure 4. Techniques for measuring mean effective stress: a) stress transient dip test technique, b) strain transient dip test technique.

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[1] If $\sigma > \sigma_i$, the dislocation sector moves in the direction of the applied stress, but when $\sigma < \sigma_i$, this sector can move opposite to the direction of the effects of this stress.

The usual procedure is to measure the dependence of relaxation $\dot{\sigma}_t$ or creep $\dot{\varepsilon}_t$ immediately following various reductions of stress and the value of the median effective stress is obtained through interpolation of the dependence $\dot{\sigma}_t$ or $\dot{\varepsilon}_t$ on $\Delta\sigma$ to $\dot{\sigma}_t = 0$ or $\dot{\varepsilon}_t = 0$; dependence of $\dot{\varepsilon}_t$ on $\Delta\sigma$ is shown in Figure 5 - curve $\dot{\varepsilon}_t$ [15-17].

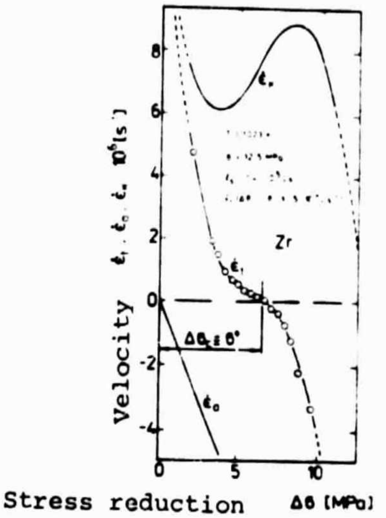

Figure 5. Experimentally determined relation between creep velocity $\dot{\varepsilon}_t$ measured immediately after reduction of stress σ by $\Delta\sigma$ and this reduction of stress [15], assumed relation [22] between initial velocity of inelastic deformation $\dot{\varepsilon}_a$ and reduction of stress $\Delta\sigma$ and relation between velocity of deformation $\dot{\varepsilon}_k$ corrected to the inelasticity effects $\dot{\varepsilon}_k$ and $\Delta\sigma$.

Implementation of the method of zero velocity of stress relaxation is considerably more demanding at high temperatures than implementation of the method of zero creep velocity. It may be for this reason that most of the so far available results of measurement of the median effective stress σ^*, or of the apparent internal stress σ_i, were obtained by the strain transient dip test technique. However, both methods yield results that, for all practical purposes, are identical. /524

A typical example of the relations between median effective stress and applied stress for various temperatures is shown in Figure 6 (Alloy Zr-2.5 Nb reinforced by precipitation [18]). From the figure it is apparent that the median effective stress measured by the strain transient dip test technique represents a significant fraction of applied stress and at the given stress it increases with increasing temperature (thermal dependence of internal stress is stronger than thermal dependence of the shear modulus).

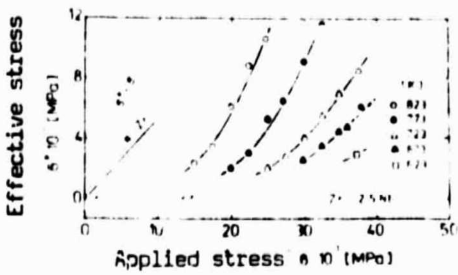

Figure 6. Relation between median effective and applied stress during stationary creep in alloy Zr-2.5 Nb reinforced by precipitation [18].

In the course of the 1970's were published many results of measurements of the median effective stress σ^* and/or internal stress /525 σ_i during creep of metals, solid solutions and alloys reinforced by precipitation. Some of these results were summed up by Takeuchi and Argon [19].

Some objections were voiced against the outlined techniques for measuring effective and/or internal stress, particularly against the strain transient dip test technique. Thus, Wilshire et al. [4-9] same as McLean et al. [20, 21] before them, and, later, some other authors (e.g. [22]) noticed an incubation period after reduction of stress $\Delta\sigma$ and came to the conclusion that after reduction of stress in a relatively wide interval of reductions of $\Delta\sigma$ the creep velocity measured immediately after stress reduction is always zero. If the stress reduction is great, according to the cited authors the creep velocity is negative due to inelasticity. Existence of the incubation period is explained by the fact that internal stress during creep is practically equal to the applied stress. Thus, results of a measurement allowing for assigning $\dot{\varepsilon}_t = 0$ to a single stress reduction $\Delta\sigma = \Delta\sigma_c$ are consequently in error.

Lloyd and McElroy [22] submitted a concept according to which inelasticity contributes to the velocity of creep measured immediately after reduction of stress σ by $\Delta\sigma$ during all reductions of stress. If the creep velocity $\dot{\varepsilon}_t$ measured immediately after reduction of stress is corrected by the contribution of inelasticity it is always positive, and it is zero only when $\Delta\sigma = \sigma$. Thus, quantities σ^* and σ_i measured by the strain transient dip test technique do not therefore make any physical sense. The eventually observed incubation period can only be the result of an approximately even velocity of forward deformation and the velocity of inelastic back deformation after reduction of stress. From Figure 5 [16, 17] it follows, however, that correction of velocity $\dot{\varepsilon}_t$ on the effect of inelasticity and/or the velocity of (reverse) inelastic deformation $\dot{\varepsilon}_a$ which shows a linear dependence on $\Delta\sigma$, leads to a velocity of $\dot{\varepsilon}_k = \dot{\varepsilon}_t - \dot{\varepsilon}_a$ which does not monotonously decrease with $\Delta\sigma$. This fact is at variance with the mentioned concept of Lloyd and McElroy [22].

Now let us describe the result of [16, 17] which strongly supports the correctness of the formulation of the strain transient dip test technique. The experiment which led to this result consisted in subjecting a sample of steel type 18Cr-10Ni to creep deformation at applied stress σ into stationary state a reduction of this stress to zero. The selected values of applied stress were relatively high to generate high values of internal stress.

After elimination of applied stress σ the velocity of deformation remained for a certain time negative--with occurrence of reverse deformation, but deformation velocity decreased in a relatively short time to zero, whereafter there set in forward deformation--the deformation velocity becoming positive. Dependence of reverse deformation on time t_R measured from the moment of elimination of applied stress is shown in Figure 7a. Points on the curves denote attainment of zero velocity of deformation. The curves represent the dependence of permanent deformation on time. To allow estimation of the extent of permanent reverse deformations--corresponding to individual applied stresses--the figure shows the corresponding elastic deformations marked by symbols ε_e. /526

The dependence of forward deformation on time t_F measured from the moment of reaching the maximum of reverse deformation or zero velocity of deformation is shown in Figure 7b. Due to the fact that forward deformation progresses during applied stress σ = 0, it can probably be explained only by the effects of internal stress [16, 17].

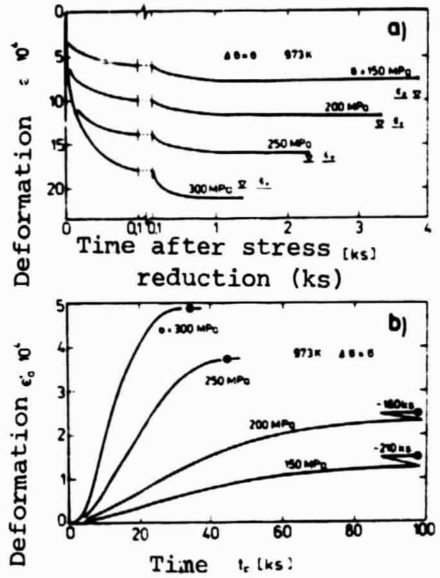

Figure 7. Dependence of reverse deformation on time t_R measured from the moment of elimination of applied stress (a); dependence of forward deformation on time t_F measured from the moment of attainment of the maximum of reverse deformation or zero velocity of deformation (b).

The deformation behavior shown in Figure 7 was observed after elimination of applied stress also in creep of zirconium [23], iron [24] and austenitic steel 16Cr-12Ni-2.5 Mo [25]. This behavior permits the conclusion that internal stress measured by the strain transient dip test technique (and also by the stress transient dip test technique) clearly makes sense physically: it represents back stress encountered by dislocation in shearing motion and the source of which is constituted by other present dislocations.

3.2 Model Concepts, Simulation of Strain Transient Dip Test Technique

As already pointed out, "actual" internal stress is a complex function of position in a crystal and the same applies to effective stress. Simultaneously with the questions prompted by effective stress there comes up quite naturally the question whether the measured apparent internal or median effective stress actually represents local internal stress or effective stress, i.e., internal and effective stress exerting their effects in microvolumes.

Thorpe and Smith [26] and later Marek [27] simulated the shearing motion of dislocation in the periodic field of internal stress. During simulation the velocity of dislocation motion was determined by random arrangement of one-dimensional local (short-range) barriers. The authors allowed for potential occurrence of forward as well as reverse thermal activation of dislocation motion and took into consideration various mechanisms controling the velocity of shearing motion. From the distribution of residence times of dislocations ahead of barriers they computed [26] the probability of occurrence of dislocation in the given position in the field of internal stress. Application of distribution of this probability to a set of noninteracting moving dislocations made it possible to simulate the strain transient dip test technique. This simulation led to the conclusion that "critical" stress--during which the median velocity of the shearing motion of dislocations is zero--represents the best estimate of the level of median (apparent) internal stress. Thus, simulation lends support to the strain transient dip test technique.

Another analysis pursuing the same objective--namely to provide an answer to the question whether effective stress measured by the strain transient dip test technique or the stress transient dip test technique can be correlated with local effective stresses active in microvolumes--was performed by Dobes [28]. This analysis based on a stochastic model of plastic deformation led to the method for simulation of the strain transient dip test technique and/or the stress transient dip test technique. The computed values of effective stress are in very good agreement with the results obtained by the strain transient dip test technique--as can be seen from Figure 8. /527
The median effective stress measured by the strain transient dip test technique corresponds well to both the median local effective stress and the weighted average of local effective stresses, provided the weighting is done by means of lengths of dislocation segments [28].

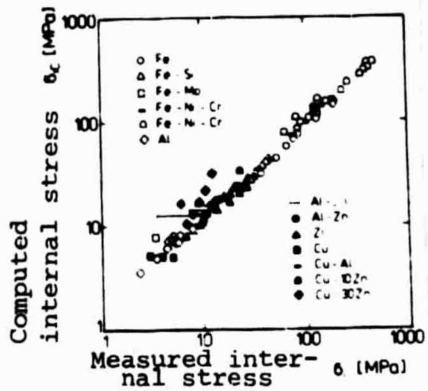

Figure 8. Correlation of theoretical values of internal stress σ_{ic} with values determined by the strain transient dip test technique σ_i [28].

Dependence of creep velocity $\dot{\varepsilon}_t$ measured immediately after reduction of stress on the extent of this reduction $\Delta\sigma$ is--as shown by the analysis--strongly affected by the density of dislocations. This is illustrated for a solid solution of Ni-4.8 Al at a temperature of 1073 K and pressure of 50 MPa in Figure 9.

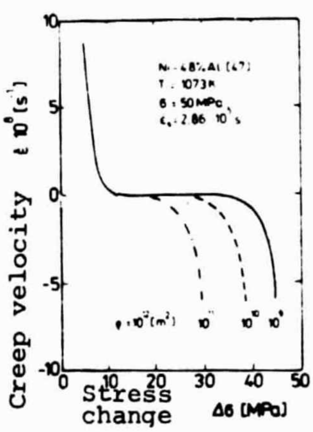

Figure 9. Dependence of creep velocity $\dot{\varepsilon}_t$ measured immediately after reduction of stress σ by $\Delta\sigma$ for varying densities of dislocations; solid solution of Ni-4.8 Al at 1073 K = 50 MPa [28].

From the figure it can be seen that when dislocation density is relatively low--$10^9 \cdot m^{-2}$--the interval of stress reductions $\Delta\sigma$ is very wide. This fact could possibly explain the above described "inconsistencies" of the results of measurement of creep velocity immediately following stress reduction; depending on the density of dislocations, on the one hand it is possible to observe the well defined stress reduction corresponded to only by the zero creep velocity $\dot{\varepsilon}_t$ and, on the other hand, a more or less wide interval of stress reductions to which can be assigned an incubation period. From Figure 9 it follows that the second of these possibilities occurs at relatively low dislocation densities. Under conditions in which it is possible to apply the strain transient dip test technique after attainment of stationary state--as borne out by numerous experimental data--the plateau of zero velocity $\dot{\varepsilon}_t$ on curve $\dot{\varepsilon}_t$ vs $\Delta\sigma$ is very short and in most cases is reduced to a point. Thus, reduction of applied stress equal to median effective stress can be defined with adequate reliability.

4. Concept of Internal and Effective Stress an Mechanical Equation of State /528

Since the measured (apparent) internal stress σ_i can be interpreted as a measure of the effect of structure--averaged for the entire sample--on the shearing motion of dislocations [3] it can be identified with the structural variable in the mechanical equation of state [2]. This equation then takes the form

$$\dot{\varepsilon} = f_{ai}\,(T\,,\ \sigma,\ \sigma_i). \qquad (3a)$$

In agreement with equation

$$\sigma = \sigma_i + \sigma^*$$

the mechanical equation state can be written also in the following form

$$\dot{\varepsilon} = f_i^*(T,\sigma,\sigma^*) = f_i^*(T,\sigma^*,\sigma_i). \qquad (3b)$$

Equation (3b) has general validity, i.e., both for transit and stationary creep. In the course of stationary creep the ratio σ_i/σ^* and, consequently, also the ratio σ^*/σ does not change. This state is attained at the moment when σ_i and, thus, also σ^* become functions of external stress and temperature:

$$\sigma_i = \sigma_i(\sigma\,,\ T);\quad \sigma^* = \sigma^*(\sigma,\ T).$$

If it follows that $(\partial\sigma_i/\partial\sigma)_T \neq 0$ and, consequently, $(\partial\sigma^*/\partial\sigma)_T \neq 0$, applied stress can be expressed as a function of σ_i and T and σ^* and T:

$$\sigma = \sigma(\sigma_i\,,T);\ \sigma = \sigma(\sigma^*\,,\ T).$$

For the stationary state then, it follows [3] that

$$\dot{\varepsilon}_s = f_a(T, \sigma) = f^*(T, \sigma^*) = f_i(T, \sigma_i). \tag{4}$$

The mechanical equation of state can thus be expressed by using temperature T and any of the stresses σ, σ^* and σ_i as independent variables. Equation (4) makes it possible to define six experimental parameters. If we choose, e.g., applied stress σ as the independent variable, we can define the apparent activation energy of creep as follows:

$$Q_c = \left[\frac{\partial \ln \dot{\varepsilon}_s}{\partial(-1/kT)}\right]_\sigma \tag{5}$$

and the parameter of sensitivity of the velocity of stationary creep to applied stress as follows:

$$m' = \left(\frac{\partial \ln \dot{\varepsilon}_s}{\partial \ln \sigma}\right)_T. \tag{6}$$

If we select effective stress as the independent variable, the definition of apparent activation energy Q^* comes out as /529

$$Q^* = \left|\frac{\partial \ln \dot{\varepsilon}_s}{\partial(-1/RT)}\right| \tag{7}$$

and definition of the parameter of sensitivity of stationary creep velocity to effective stress as follows:

$$m'^* = \left(\frac{\partial \ln \dot{\varepsilon}_s}{\partial \ln \sigma^*}\right). \tag{8}$$

Of course, internal stress σ_i can also be selected as an independent variable to define experimental parameters Q_i and m_i'. However, the physical sense of parameters Q^* and σ^* and/or Q_i and σ_i need not be always precisely clear. In certain cases (e.g., [29]) these parameters must be accepted only as phenomenological quantities which have no direct relation to the mechanism controlling creep velocity.

5. Conclusion

Discussion is offered of the concept that deformation of metallic materials at high temperatures is not due to the effects of the entire applied stress σ, but only by its part represented by the difference between $\sigma - \sigma_B$, where σ_B is back stress characterizing resistance of the material against plastic deformation.

Back stress tends to be identified with internal stress σ_i or with "friction" stress σ_0. Internal and friction stress depends in general on applied stress and on temperature. Both internal and friction stress can be determined experimentally. In addition, internal stress can be deduced from model concepts. The measuring techniques and model concepts are briefly described.

If back stress σ_B does not depend on applied stress, it can be identified as threshold stress σ_T below which dislocation creep does not take place. The thermal dependence of threshold stress derives exclusively from the thermal dependence of the shearing modulus of elasticity.

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