Creep and Fracture of Engineering Materials and Structures

Edited by:

B. Wilshire
Department of Metallurgy and Materials Technology, University College, Swansea.

D. R. J. Owen
Department of Civil Engineering, University College, Swansea.

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PREFACE

The response to the Call-for-Papers from authors throughout the world has indicated the widespread interest in the subject of 'Creep and Fracture of Engineering Materials and Structures'. The papers accepted for publication in the Proceedings have been separated into six sections covering the major areas of interest, namely, creep mechanisms, deformation processes in particle-strengthened alloys, creep fracture processes, creep and fracture of ceramics, materials behaviour at elevated temperatures and the design and performance of components and structures.

The Proceedings are printed from direct lithographs of authors manuscripts and the editors cannot accept responsibility for any inaccuracies, comments or opinions expressed in the papers. However, the organisers wish to thank the authors for presenting their work and ideas in the context of the overall position currently reached in the area relevant to the theme considered. In this way, the Conference provides both an overview of the different approaches being developed in various centres active in the field of creep and fracture and an indication of the principal avenues along which future activities should be directed.

The organisers wish to acknowledge the generous sponsorship of the Conference provided by the United States Air Force, European Office of Aerospace Research and Development and the Department of the Navy, Office of Naval Research. The sponsorship received from Mand Testing Machines Ltd., Eurotherm Ltd. and Automatic System Laboratories Ltd. towards the social programme of the Conference is also gratefully acknowledged.

B. WILSHIRE D.R.J. OWEN

Swansea, March 1981.

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DISLOCATION CREEP IN SUBGRAIN-FORMING PURE METALS AND ALLOYS

A. S. Argon, F. Prinz, and W. C. Moffatt

Massachusetts Institute of Technology
Cambridge, Massachusetts 02139, U.S.A.

SUMMARY

Steady state creep and structural transients are considered in subgrain-forming pure metals and alloys. The formation of subgrain boundaries by an evolutionary clustering process is described, together with the process of subgrain boundary annihilation by coalescence, due to their migration, leading to a structural steady state. The role of the various components of the dislocation content of sub-boundaries is discussed with particular emphasis on the debris content that plays a key role in the mobility of sub-boundaries under stress. Finally, internal stresses, measurable by the stress dip test in such metals, are attributed to the flexing under stress of sub-boundaries. The calculated magnitudes of such internal stresses are in very good agreement with reported measurements. These internal stresses are responsible for changing the normal third power law of the strain rate stress relation into one of higher power.

1. INTRODUCTION

Creep deformation at high temperatures in engineering structures is of unquestioned technological importance. In its relatively idealized form in pure metals and single phase alloys, creep has been widely investigated. The phenomenology of creep and the many attempts to explain it mechanistically have been reviewed recently [1-4]. In spite of the fact that considerable detailed criticism has been directed against it, the simple picture of Bailey [5] and Orowan [6] that high temperature creep combines two competing mechanisms of strain hardening and thermal recovery has found wide acceptance and forms the basis of nearly all current research on the subject. A point of particular interest has been steady state creep in pure metals and Class II alloys, its specific functional form, and its basis in fundamental processes of the glide, climb, and clustering of dislocations into subgrain walls. This will also be the subject of primary interest to us in this communication, where we will attribute the high stress exponent of the steady state creep rate to long range dynamic internal stresses, and will propose the bowing of subgrain walls under stress as the principal source of this internal stress. Furthermore, we will consider the mobility of subgrain walls under stress as an important part of steady state and present an outline development for the production and annihilation of subgrain walls. We will then proceed and furnish some new experimental observations on the specific dislocation content of sub-boundaries in creeping alloys and discuss the role of this dislocation structure in governing the mobility of sub-boundaries. In all of this we will strive for internal consistency but leave the more complete mechanistic description of the complex evolutionary processes of steady state to a future communication.

2. STEADY STATE CREEP

2.1 Internal Stresses

The analyses by Bird, Mukherjee, and Dorn [1], Mohamed and Langdon [7], and others [2-4] of many creep studies have established that the steady state creep rate is given by an expression of the form

$$\dot{\varepsilon} = A \left(\frac{\mu\Omega}{kT}\right)\left(\frac{\chi}{\mu b}\right)^3\left(\frac{D}{b^2}\right)\left(\frac{\sigma}{\mu}\right)^m \tag{1}$$

where b , μ , Ω , χ , and D are the interatomic distance, the shear modulus, the atomic volume, the stacking fault energy and the self diffusion constant, and where m , the stress exponent, is often in the range of 5 . Argon and Takeuchi [8] have shown that a creep rate expression of the above form but with an exponent of $m = 3$ is a natural result of a steady state sequence of processes of gliding and climbing of dislocations. Under a tensile stress σ , the process starts by the generation of dislocations at sources after the passage of a characteristic waiting time requiring the climb of a short segment a critical distance $\lambda \simeq (0.1)\mu b/\sigma$, to become free from surrounding dislocations. The generated dislocations move by a predominantly glide motion over a storage distance L that is often larger than subgrain dimensions, where they become trapped by other dislocations within a trapping distance w . There they are eventually annihilated after some further climb toward the trapping dislocations of opposite sign, resulting in a mobile dislocation density of

$$\rho_m = \frac{C_m}{b^2}\left(\frac{w}{L}\right)\left(\frac{\sigma}{\mu}\right)^2\left(\frac{\chi}{\mu b}\right) \quad , \tag{2}$$

where C_m is a constant of order 500 and χ the stacking fault energy. The model considers that most of the time, τ_a, is spent by a dislocation in the various climb steps leading to eventual annihilation in which the velocity v_c of the dislocation is given by [9]

$$v_c = v_o \frac{\sigma}{\mu} = C_c \frac{D}{b}\left(\frac{\mu\Omega}{kT}\right) c_j \left(\frac{\chi}{\mu b}\right)^2\left(\frac{\sigma}{\mu}\right) \quad , \tag{3}$$

where C_c is a numerical constant of order 10^3, σ is the climb producing normal stress acting across the half plane of the dislocation, and c_j is the jog concentration along the extended dislocation. This leads to an average velocity $\bar{v}$ during the life time of a dislocation:

$$\bar{v} = L/\tau_a \quad ; \quad \tau_a = \pi(1-\nu)w^2/2bv_o \quad ; \quad w = \mu b/4\pi\sigma \ . \tag{4a,b,c}$$

Apart from the specific dependence of the strain rate on the stacking fault energy given by Eqn.(1), which results in part from the form of the climb velocity of Eqn.(2) and in part from a factor $(\chi/\mu b)$ given in Eqn.(2) that gives a decreased source configuration probability discussed by Argon and Takeuchi [8], the form of Eqn.(1) with $m = 3$ is identical to the one introduced by Weertman [10] as the rational creep law.

Argon and Takeuchi [8] have proposed further that powers higher than 3 in the stress exponent of Eqn.(1) in subgrain forming alloys are due to dynamic internal stresses that result from the bowing of the boundaries of such subgrains as they migrate under stress. According to this model, most sub-boundaries bow out under an applied stress to an amplitude y over the sub-boundary facet length of δ, as the sub-boundary nodes, subject to geometrical constraints, drag behind as shown in Fig. 1. Such flexing of sub-boundary

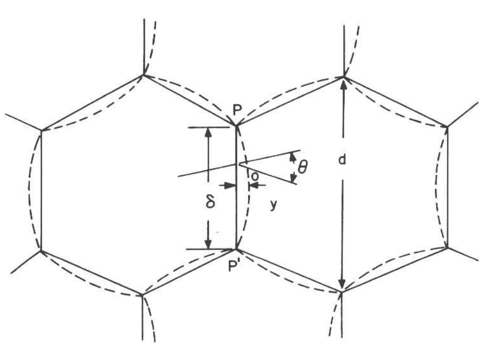

Fig. 1 Idealized hexagonal subgrains with randomly flexed sub-boundaries under stress.

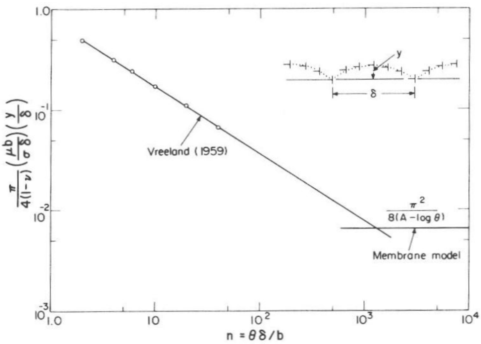

Fig. 2 Amplitude of sub-boundary flexure under stress for sub-boundaries containing small numbers of dislocations, shown by the slanted line (from Argon and Takeuchi [8]).

facets, treated as a collection of discrete dislocations, has been studied by Vreeland [11], who finds, as shown in Fig. 2, that the amplitude of flexing of a sub-boundary with lattice misorientation θ under an applied stress σ is given by

$$\frac{y}{\delta} = 0.8 \frac{4(1-\nu)}{\pi}\left(\frac{\sigma\delta}{\mu b}\right)\left(\frac{b}{\theta\delta}\right)^{2/3} . \tag{5}$$

It is well known that when the plane of a low angle boundary is rotated from its lowest energy orientation, large and long range internal stresses are produced [12]. On this basis, Argon and Takeuchi [8] have associated the long range stresses measured in the stress dip experiments [13] to such rotations of sub-boundary planes and proceeded to calculate the amplitude σ_i of these randomly positive and negative stresses by treating the volume enclosed by the flexed lobes of sub-boundaries as if they were partially constrained shear transformations with transformation shear strains of θ. This has given

$$\sigma_i = (1/8)(\mu\theta)(y/\delta) , \tag{6}$$

which upon substitution of Eqn.(5) and the use of the steady state relationship between subgrain size d and stress

$$d = 2\delta = K(\mu b/\sigma) \tag{7}$$

$$\sigma_i/\mu = (C_i(1-\nu)/\pi)(K\theta)^{1/3}(\sigma/\mu)^{2/3}; \quad (C_i = 0.317). \tag{8}$$

Takeuchi and Argon [3] have shown from the analysis of creep experiments of others that K varies from near 10 for close packed metals to near 60-70 for alkali halides and oxides, with an overall average being near 30. This dependence of the internal stress σ_i on the applies stress σ is plotted

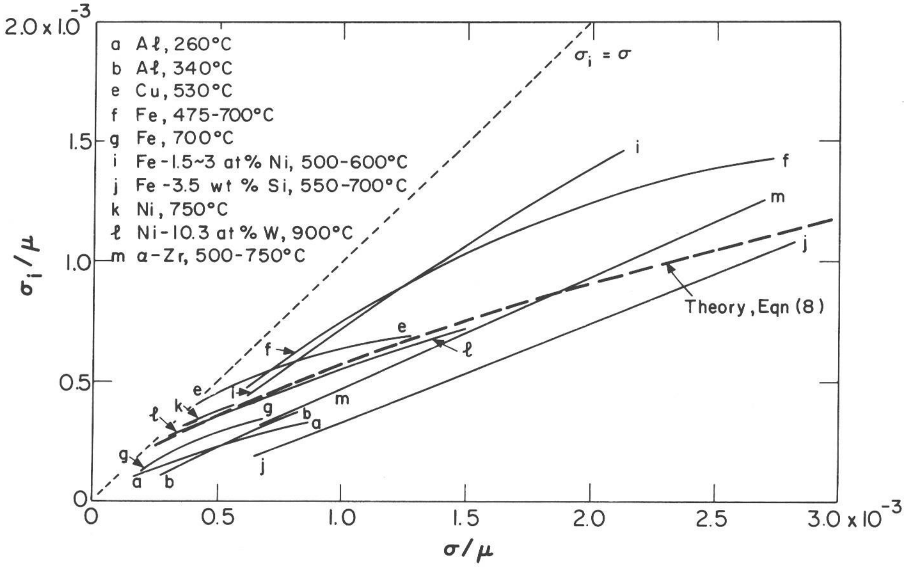

Fig. 3 Comparison of computed dependence of internal stress on applied stress, with the measured dependence for a variety of pure metals and Class II alloys (from Argon and Takeuchi [8]).

in Fig. 3 together with the internal stresses measured with the stress dip test in many pure metals and Class II alloys for a typical lattice rotation of $\theta = 1°$. The agreement between theory and the entire family of experimental results is good.

If the stress in the general expression for steady state creep given by Eqn.(1) with an exponent $m = 3$ is now interpreted to be the effective stress

$$\sigma_e = \sigma - \sigma_i \quad , \tag{9}$$

the overall creep rate should then be given by an expression of

$$\dot{\varepsilon} = A \left(\frac{\mu\Omega}{kT}\right)\left(\frac{\chi}{\mu b}\right)^3\left(\frac{D}{b^2}\right)\left(\frac{\sigma}{\mu} - \frac{C_i\,(1-\nu)}{\pi}\,(K\theta)^{1/3}\left(\frac{\sigma}{\mu}\right)^{2/3}\right)^3 \tag{10}$$

where A is a numerical constant of the order of $(6\mathrm{x}10^6)c_j$ [8]. The stress dependence of this equation for $K = 30$ and $\theta = 1°$ is plotted in Fig. 4 over a decade of σ/μ in which most power-law creep experiments are conducted. Clearly, as Eqn.(10) and the plot of Fig. 4 shows, $\dot{\varepsilon}$ exhibits a threshold behavior at a stress that makes the content of the last parenthesis vanish. We dismiss this behavior as unreal and expect that both grain-boundary sliding and Nabarro-Herring creep will add a substantial component of strain rate to the total in this low stress range. That this must be the correct explanation is shown from much of the actual internal stress measurements shown in Fig. 3 which indicate that at these low stress levels the internal stress makes up almost the entire applied stress leaving only the mechanisms that do not involve dislocation mobility to produce inelastic strain. The plot of Fig. 4 shows that in much of the useful range of stress $(10^{-4}-10^{-3})\mu$ the apparent stress exponent is in the neighborhood of 5 , going from 6 to 4.5. This is the normal reported behavior. Much above a stress of $2\times10^{-3}\mu$, the curve goes to an asymptotic form of a power of 3 as $\sigma_i/\sigma \to 0$. In this range, however, additional low temperature processes involving thermally activated overcoming of slip obstacles can produce increasingly large components of strain rate which have been excluded from the model that gives rise to Eqn. (9). It is this additional component that results in the so-called power-law break-down behavior.

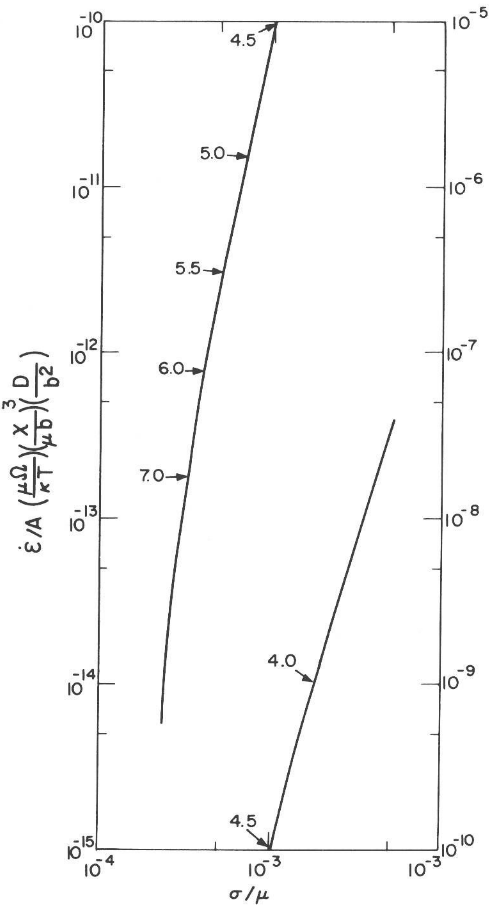

Fig. 4 Computed dependence of steady state creep strain rate on the applied stress. Numbers indicate the level of the local power law creep exponent (from Argon and Takeuchi [8]).