

Creep and Fracture of Engineering Materials and Structures

Edited by:

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Creep and Fracture of Engineering Materials and Structures

PREFACE

The International Conference held during 1st-6th April, 1990, was the Fourth in the Series organized at Swansea on the theme of 'Creep and Fracture of Engineering Materials and Structures'. As with the previous conferences in 1981, 1984 and 1987, authors from virtually every centre active in the field throughout the world have contributed to make these Proceedings a comprehensive 'state-of-the-art' coverage of recent theoretical and practical advances in creep and creep fracture.

In addition to attracting authors of international stature, the present Conference Series has always been most fortunate in its sponsors. For their support of the social evenings which invariably provide such a splendid forum for informal discussions, we wish to express our gratitude to the sponsors of the 1990 Swansea Conference, namely

ESH Testing Limited, Brierley Hill, West Midlands
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The social programme was also greatly enhanced by a Civic Reception and a concert from the Morrision Orpheus Choir which was organized at the Brangwyn Hall by the Lord Mayor and the Swansea City Council. For their generosity and hospitality we offer our thanks on behalf of all delegates.

A successful Conference Series also depends on many people contributing well beyond the call of duty. The 1990 Conference was again held, with the wholehearted endorsement of the delegates attending the earlier meetings, at the Clyne Castle site of the University College, Swansea. We are again indebted to the staff at Clyne Castle for their outstanding service which contributed so much to the splendid atmosphere generated. Moreover, for many months of effort which demonstrated their administrative talents, patience and unfailing cheerfulness, we offer our genuine appreciation to the Conference Secretaries, Mrs. Janice Price and Mrs. Vivienne Jenkins.

B. Wilshire
R.W. Evans
University College Swansea
April, 1990.

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ON THE EVIDENCE FOR DIFFUSIONAL CREEP PROCESSES

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SUMMARY

The few inconclusive experimental observations quoted as evidence for diffusional creep processes do not justify the common assumption that high temperature creep of polycrystalline materials normally occurs at low stresses by stress-directed vacancy flow without dislocation movement.

INTRODUCTION

The high temperature creep properties of crystalline solids have generally been discussed by reference to the variations of the secondary or steady-state creep rate ($\dot{\epsilon}_s$) with stress (σ), grain size (d) and temperature (T), using a power law relationship of the form

$$\dot{\epsilon}_s = A \sigma^n (1/d)^m \exp -Q_c/RT \quad (1)$$

where A , n and m are constants and Q_c is the activation energy for creep. However, when this approach has been adopted, the values of the 'constants' n , m and Q_c have usually been found to vary depending on the creep conditions imposed. The observed variations in n , m and Q_c have then been explained on the basis that different mechanisms control the creep behaviour in different stress/temperature regimes. Furthermore, it has also been widely assumed that the dominant creep mechanism in any stress/temperature regime can be identified by comparing the measured values of n , m and Q_c with the values predicted theoretically for individual creep processes.

During the last decade, a direct challenge to these traditional 'steady-state' or 'power-law' ideas has been provided by an alternative theoretical and practical approach to creep, termed the θ Projection Concept (1,2). In this case, the view that a 'steady-state' creep rate occurs is abandoned. Instead, the normal creep curves observed at high temperatures are considered in terms a decaying primary curve and an accelerating tertiary component, ie the 'secondary stage' is merely the period of apparently constant creep rate found when the decaying primary creep rate is offset by the acceleration due to tertiary processes, eg intergranular damage accumulation,

microstructural instability, etc. The θ Project Concept then offers detailed equations which accurately describe the full creep curve shape and the changes in creep curve shape with changing stress and temperature (2).

A firm theoretical foundation has been provided for this new approach, based on the micromechanisms responsible for primary and tertiary creep (2). Moreover, the θ Projection Concept offers the major practical advantage of allowing the long-term creep and creep rupture properties of metals and alloys to be predicted accurately by extrapolation of data derived from short-term constant-stress creep curves (1,2). Thus, analysis of creep curves obtained only at high stresses for polycrystalline copper (3) predicts exactly the gradual decrease in n and Q_c with decreasing stress and temperature which has traditionally been taken as evidence for a transition from dislocation to diffusional creep. Indeed, as evident from Figure 1, analysis of data recorded at high stresses predicts all features of the deformation mechanism map constructed for polycrystalline copper (4). On this basis, no changes in creep mechanism need be invoked to explain the gradual decrease in n and Q_c with decreasing stress and temperature. Instead, the complex stress and temperature dependences of n and Q_c are seen to be simple consequences of the variation in creep curve shape with test conditions which can now be described quantitatively using the θ Projection Concept (1,2,3).

The explanation for the changes in n and Q_c provided by the θ Projection Concept appear to conflict directly with the traditional view that variations in n and Q_c are attributable to a change from dislocation to diffusional creep processes. Yet, diffusional creep theories are elegant (5-7) and have been widely assumed to account for the creep properties expected at low stresses. For these reasons, it is useful to review the experimental data which has been quoted as evidence for diffusional creep.

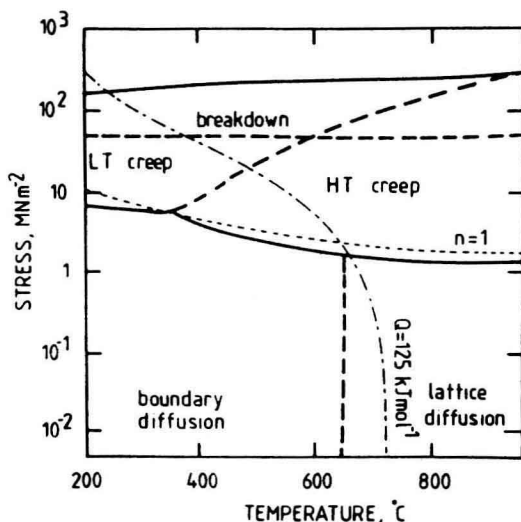


Figure 1. Deformation mechanism map for polycrystalline copper (4), together with contours of $n \sim 1$ and $Q_c \sim 125 \text{ kJmol}^{-1}$ obtained by analyses of high-stress creep data using the θ Projection Concept (3).

THE STRESS AND TEMPERATURE DEPENDENCE OF CREEP AT LOW STRESSES

The general acceptance of diffusional creep concepts is such that, since the theories predict that $n \sim 1$ (5-7), creep measurements showing n values close to unity are commonly taken as sufficient reason for assuming that diffusional creep processes are rate-controlling. Yet, there is unambiguous evidence to prove that this tautological argument need not be valid. For instance, since the original work of Harper and Dorn (8), many studies have shown that the n value decreases towards unity with decreasing stress for both single and polycrystalline materials, as illustrated for aluminium in Figure 2 (9). Diffusional creep processes, which would be imperceptibly slow with single crystals, cannot be relevant when n values close to unity occur with single crystals at low stresses. Thus, n values near to unity do not guarantee that diffusional creep processes are rate-determining.

For single crystals of materials such as aluminium, creep must occur by dislocation creep processes even at low stresses. The observation that single crystals display creep rates of the order of those found for polycrystalline materials (Figure 2) then suggests that creep of both single and polycrystalline materials occurs by diffusion-controlled generation and movement of dislocations at both high and low stresses, (8,9). In this way, a straight forward explanation is also provided for the observation that Q_c decreases from values around those for self diffusion as the temperature is decreased in the $n \sim 1$ regime. This transition has usually been explained on the basis that, in the diffusional creep regime, diffusion occurs predominantly through the lattice at high temperatures (known as Nabarro-Herring creep) and preferentially along grain boundaries at lower temperatures (termed Coble creep). However, a similar transition in Q_c value occurs even when creep unquestionably takes place by dislocation processes, a fact attributed to pipe-diffusion becoming progressively more important as the temperature is decreased towards $0.4T_m$ (where T_m is the absolute melting point). Hence, n and Q_c measurements do not prove whether dislocation or diffusional creep mechanisms are dominant.

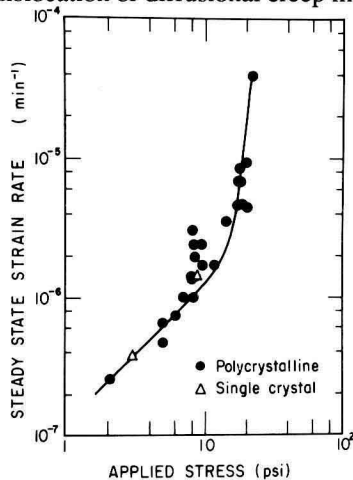


Figure 2. Stress dependence of the secondary creep rate for aluminium at temperatures between 914 and 925K (9).

THE GRAIN SIZE DEPENDENCE OF CREEP AT LOW STRESSES

The results in Figure 2 show that for aluminium (8,9), as with many other single-phase materials (10,11,12), single crystals and polycrystalline samples display creep rates of similar magnitude under the same low-stress test conditions. In relation to equation 1, this means that the creep rate is almost grain size insensitive, so that $m \sim 0$. Values of m close to zero would be expected when dislocation processes control creep behaviour, a conclusion compatible with the proposal that creep of aluminium occurs by diffusion-controlled generation and movement of dislocations in both the $n \sim 5$ and $n \sim 1$ regimes of Figure 2. In contrast, diffusional creep theories predict $m = 2$ for Nabarro-Herring creep and $m = 3$ for Coble creep (5-7). Data consistent with these large m values have been reported, as shown for copper in Figure 3 (13). However, there are problems associated with interpretation of creep data obtained for pure metals at high fractions of their melting points.

With pure polycrystalline copper tested at temperatures from about 0.4 to 0.6 T_m , a minimum creep rate occurs when the decaying primary rate is offset by the acceleration due to intergranular damage development (3). However, at higher temperatures, the decay in the primary creep rate and certainly the acceleration during the tertiary stage can be modified considerably by grain growth and recrystallization, giving minimum creep rates higher than those expected for microstructurally-stable material. Indeed, at temperatures between 0.6 and 0.8 T_m depending on the purity of the polycrystalline copper (14), dynamic recrystallization can suppress grain boundary cavitation so that fracture occurs by necking to a point (rupture) rather than by intergranular crack development. Under these conditions, it would be anticipated that the effects of microstructural instability would be less pronounced under fixed test conditions when longer annealing times and/or higher annealing temperatures are used to produce larger more stable grain sizes. Consequently, considerable

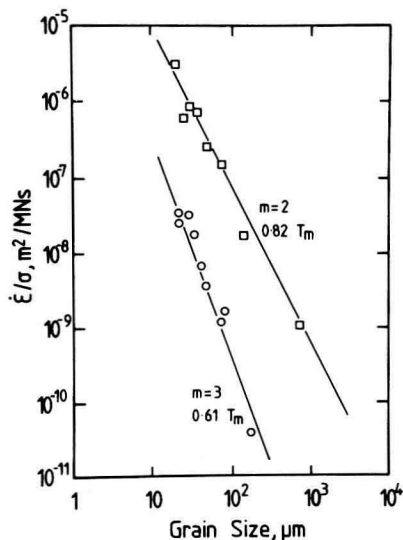


Figure 3. The variation of the creep rate per unit stress with grain size for polycrystalline copper tested at 0.61 and 0.82 T_m (13).