MICROMECHANISMS OF PLASTICITY AND FRACTURE

Editors

M.H. Lewis
D.M.R. Taplin

SM Publications
University of Waterloo Press

Parsons Press, Trinity College
Inaugural Volume, 1983

Solid Mechanics Division
University of Waterloo
Waterloo, Ontario, Canada

Parsons Press
Trinity College
Dublin, Ireland

MICROMECHANISMS OF PLASTICITY AND FRACTURE

Editors:

M.H. Lewis
Reader in Materials Physics
Warwick University, England

D.M.R. Taplin
Professor of Engineering Science
Trinity College, Dublin

University of Waterloo Press

Parsons Press, Trinity College
Inaugural Volume, 1983

ISBN: 0-88898-045-0

INTRODUCTION

This volume collects together a series of review papers most of which were originally presented at a symposium held at St. Catherine's College, Oxford, in honour of Dr. J.W. Martin. The papers cover aspects of the micromechanisms of strength and fracture of engineering materials. As a group, the papers provide a useful basis for a senior undergraduate or graduate course at universities, and they have been successfully used in this way for courses in the Department of Mechanical Engineering at the University of Waterloo, the Department of Mechanical and Manufacturing Engineering, Trinity College, Dublin and the Department of Physics, University of Warwick. The volume is also designed for researchers in the field of micromechanisms and for others who require a survey of the current understanding of plasticity and fracture mechanisms encompassing a range of engineering materials. The papers are written by authors who have a recognised expertise in their own field of research. The coverage is by no means complete and the volume parallels more general texts by providing detailed and up to date studies of some important areas of development in materials physics.

A common philosophy in the research described in these papers, and that which epitomises our own view of the subject area of "Materials Science and Engineering", is the emphasis on *real* engineering solids as opposed to structurally simplified *models* and upon *real* service conditions in engineering practice. The model material approach probably received undue attention during the 1950's and 1960's with much work, for example, on pure single crystals. Single crystals provide rather poor models for the analysis of deformation and fracture processes in even the simplest engineering materials. Many researchers have subsequently attempted to correct the balance of research effort with attention to realistic materials and the solution of the complex problems which they present. An excellent example is the work at N.P.L. which has made impressive progress in analysing and solving a number of failure problems associated with the grain-boundary segregation of impurities. Such problems are of enormous technological and economic importance but their solution is scientifically challenging and in the latter example necessitates the use of sophisticated techniques of electron spectroscopy.

This attitude to materials research has been pursued consistently at Oxford by John Martin over the past twenty-five years and thereafter by many of his former graduate students who are the authors of a number of the papers in this volume. John Martin, Sc.D., Ph.D., D.Phil., M.A., F.I.M., C.Eng., Goldsmiths' Fellow, St. Catherine's College and University Lecturer in the Department of Metallurgy and Science of Materials at Oxford University since 1957, has made many significant contributions to the understanding of micromechanisms of strength and fracture and he continues to be very active as a researcher today. Apart from his many papers, he has published several well-received books, the latest being *Micromechanisms in Particle Hardened Alloys*. In the present volume we recognize his contribution and look forward to producing a further volume in this vein in another decade.

This book is also the inaugural volume of the Parsons Press of Trinity College, Dublin, established to publish books of interest to mechanical, materials and manufacturing engineering. For this first volume the Parsons Press is co-publishing with Solid Mechanics Publications of the University of Waterloo, Canada and it is a special pleasure to record our appreciation to Dr. Donald E. Grierson, Managing Editor, Mrs. Linda Strouth (who prepared the complete manuscript) and Mr. David Bartholomew, who designed the graphics, including the new Parsons Press logo.

M.H. Lewis, D.Phil.
Reader in Materials Physics
University of Warwick

D.M.R. Taplin, D.Phil.
Professor of Engineering Science
Trinity College, Dublin

June, 1983

CONTENTS

MICROMECHANISMS OF PLASTICITY AND FRACTURE
SM Publications
University of Waterloo Press

INAUGURAL VOLUME
Parsons Press, 1983
Trinity College, Dublin

THE INTERACTION OF DISLOCATIONS WITH HARD PARTICLES

F.J. Humphreys

Department of Metallurgy and Materials Science
Royal School of Mines
Imperial College of Science and Technology
London SW7 2BP, England

1.1 INTRODUCTION

The mechanical behaviour of a metal may be changed drastically by the presence of a fine dispersion of strong second phase particles. In particular, the yield stress and work hardening rates at low temperatures are raised; and at high temperatures where many other forms of strengthening become less effective, the particles, if they do not dissolve or coarsen, are still good barriers to dislocation movement. In addition, the particles may inhibit primary recrystallisation, thereby retaining the strengthening effect of a dislocation substructure to high temperatures.

Because the important dislocation interactions are localized at well defined sites, i.e., the particles, a study of these interactions is more amenable to experimental investigation than is often the case for single phase materials. It is now possible at least in simple cases, to explain some important aspects of the mechanical properties of two-phase alloys in terms of these interactions.

In this paper, we shall be concerned mainly with particles which do not deform when the matrix is deformed, and with low temperature behaviour. Above all, this paper is concerned with the mechanisms of dislocation/particle interactions, and no detailed account will be given of theories of the yield or work hardening of these materials.

1.2 YIELD

1.2.1 *The Mechanism of Yield*

For alloys containing weak particles, the yield behaviour depends on the nature of the short range interactions between dislocations and particles - see for example the review of Brown and Ham [1.1]. However for strong particles the Orowan mechanism as shown in Figure 1.1 operates.

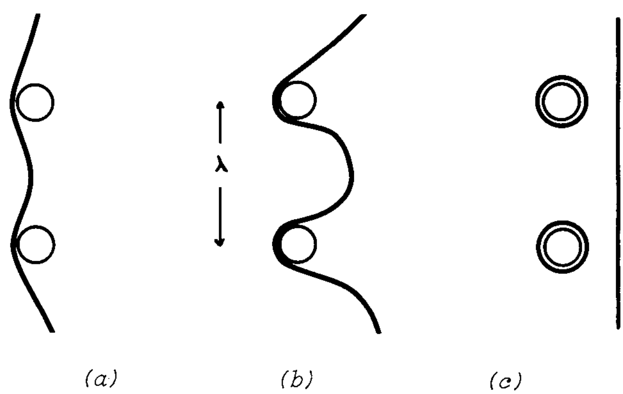

Figure 1.1 - The Orowan Mechanism

The possibility of the screw component of the dislocation cross-slipping before an Orowan loop is formed has been considered [1.2], but, as will be discussed later, this is now considered unlikely unless there are considerable misfit stresses associated with the particle. Even if this were to occur, then it is unlikely to affect the yield stress as various authors have demonstrated [1.1,3,4].

The Interaction of Dislocations

It is to be expected that two-phase alloys will not show a sharp yield point, but will exhibit a significant amount of microstrain before general yield occurs. This is because the distribution of particles is not uniform, and dislocations will propagate through the material in stages as the stress is raised, finding the easy paths first [1.5,6,7].

Consideration should be given to the possibility that even strong oxide particles may deform before the Orowan stress is reached. At the critical part of the Orowan process the stress at the particle interface is (Figure 1.1(b)), $\sim \frac{Gb}{2r}$, where G is the shear modulus of the matrix, b the Burgers vector, and r the particle radius. If this is equal to the strength of the particle, which can be taken to be a maximum of $\sim \frac{G'}{10}$ where G' is the shear modulus of the particle, then the particle will deform, i.e.,

$$r < 5\,\frac{Gb}{G'}\ . \qquad (1.1)$$

A calculation based on energy considerations [1.8] yields a result of similar magnitude. For many oxide particles where $G' > G$, this critical radius is very small; nevertheless, there is some evidence from measurement of yield stresses to suggest that alumina particles of $r < 75\text{Å}$ in copper [1.9] and $r < 25\text{Å}$ in silver [1.10] deform at yield.

1.2.2 *The Yield Stress*

Comparison of experimental results with theory leaves little doubt that oxide hardened alloys generally yield at the Orowan stress. The simplest form of this is:

$$\tau_y \sim \frac{Gb}{\lambda}\ . \qquad (1.2)$$

The calculation of the Orowan stress has been considerably refined during the past 15 years, and detailed discussions may be found in the literature [1.1,4,6,11,12]. However, equation (1.2) still provides a useful approximate value.

If the particle-free matrix has a significant yield stress due for example to solution or order hardening, or if several types of particle are present, then consideration must be given to the ways in which these separate hardening mechanisms contribute to the yield stress [1.1,5,6,7]. If there are two sets of obstacles with greatly differing strengths, e.g., solute atoms and particles, which would individually give yield strengths of τ_1 and τ_2, then the resultant yield strength, τ_y, is given by:

$$\tau_y = \tau_1 + \tau_2 \ . \tag{1.3}$$

However, for obstacles of similar strength, e.g., two types of precipitate, or precipitate plus forest dislocations, then:

$$\tau_y^2 = \tau_1^2 + \tau_2^2 \ . \tag{1.4}$$

1.3 THE OBSERVED MECHANICAL BEHAVIOUR

It is not the intention in the paper to give an exhaustive review of the mechanical properties of dispersion hardened alloys. However, before discussing in detail the mechanisms of interaction of dislocations and particles, a summary will be given of the various phenomena which have been observed during mechanical testing, particularly where they differ from those observed during the deformation of single phase metals.

1.3.1 *The Work Hardening of Polycrystals*

Stress-strain curves of some internally oxidised copper polycrystals are shown in Figure 1.2. Comparison with pure copper shows that although the work hardening rates of the two-phase alloys are greater initially, after a few per cent strain, the curves become parallel, the difference in the flow stresses being approximately the amount by which the yield stresses differ. The temperature dependence of the flow stress in such alloys [1.13,14] is also found to be similar to pure copper. Measurements of stored energy after high strain deformation [1.15,16] again indicate that the values for dispersion hardened copper are comparable to those for pure copper. Thus for polycrystals, we conclude that particles exert their greatest effect on the mechanical properties during the earliest stages of deformation.

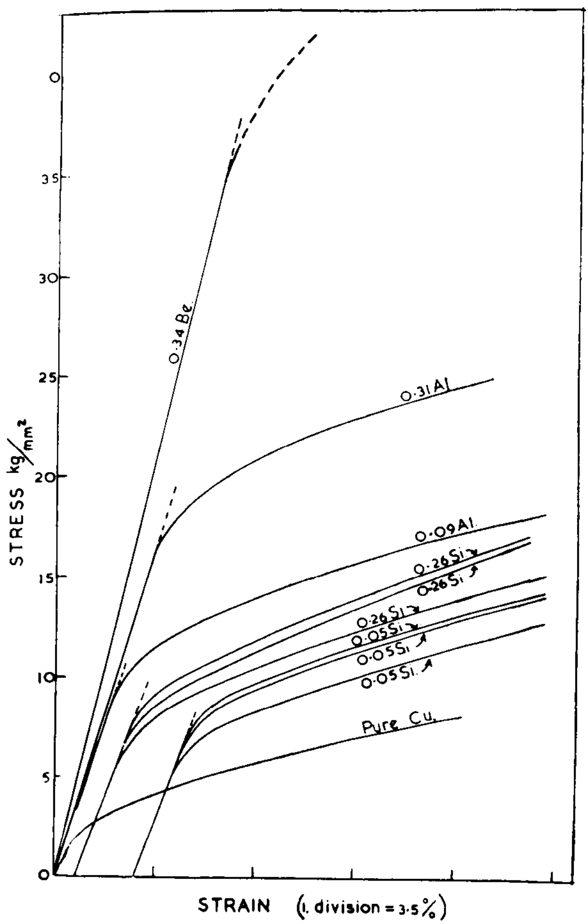

Figure 1.2 - Stress-Strain Curves for Internally Oxidised Copper Polycrystals [1.13]

1.3.2 *The Work Hardening of Single Crystals*

The effect of nondeformable particles on the stress-strain curves of copper single crystals is now well documented, and examples are shown in [1.3,7,17,18,19], Figure 1.3. It can be seen that an increased volume fraction of particles increases the initial work hardening rate, and changes in the shape of the curve, transforming the linear stage I easy glide region of copper into a stage of parabolic hardening.

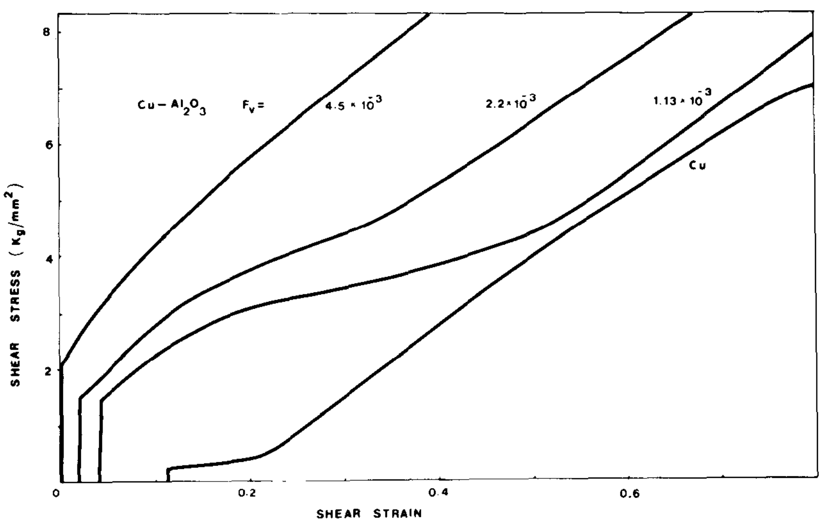

Figure 1.3 - Stress-Strain Curves for Single Crystals of Cu-Al_2O_3

The particles do not greatly affect the later stages of work hardening as can be seen for example in Figure 1.3. Electron microscopy [1.3] has shown that the deformation in the matrix during the initial parabolic stage of hardening is by movement of primary dislocations, and it is therefore reasonable to compare this stage with stage I of single-phase crystals. Ebeling and Ashby [1.17] found that the flow stress (τ) of Cu-SiO_2 crystals in this region was given by the relationship:

The Interaction of Dislocations

$$\tau = \tau_y + K\left(\frac{\gamma F_v}{r}\right)^{1/2} , \qquad (1.5)$$

(γ = shear strain, F_v = volume fraction, and K = constant).

Subsequent experiments have shown this equation to be valid to a first approximation for a variety of oxide hardened crystals. However, detailed analysis of the stress-strain curves shows that, particularly for particles of d < 250Å, the parabolic hardening is preceded by a short region of low hardening rate [1.7, 18,19,20], and that at large strains ($\gamma \gtrsim 0.1$) the work hardening may fall substantially below the parabola of equation (1.5), [1.20].

1.3.3 *The Bauschinger Effect*

Both single crystals and polycrystals of alloys containing non-deformable particles exhibit a large Bauschinger effect [1.21-1.26]. As shown in Figure 1.4 the magnitude, measured as the difference between the forward and reverse yield stresses, increases with prestrain, reaching a maximum at $\gamma \sim 0.05$.

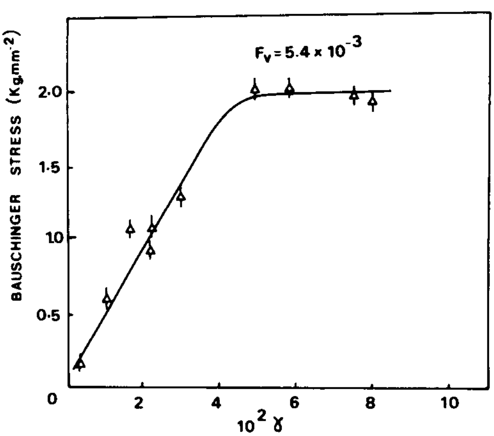

Figure 1.4 - The Effect of Prestrain (γ) on the Magnitude of the Bauschinger Effect in Cu-Al_2O_3 [1.24]

The effect has been measured in copper crystals containing dispersions of SiO_2, Al_2O_3 and BeO, and for particles of diameter less than ∿ 1000Å there is little indication that the Bauschinger effect is sensitive to particle size or shape. In oxide strengthened copper crystals the Bauschinger effect becomes very small at temperatures above ∿ 350K and is much reduced if a crystal is 'annealed' at room temperature, [1.24,25].

1.3.4 *Recovery*

Deformed single crystals of copper containing oxide particles show a pronounced recovery effect at ambient temperatures [1.18,18,24,25, 27]. If a crystal is deformed at 77K, unloaded and left at room temperature for a few hours, and then retested at 77K, a drop in flow stress, often a large fraction of the work hardening is found. The amount of recovery is dependent upon the prestrain as shown in Figure 1.5 reaching a maximum at γ ∿ 0.05.

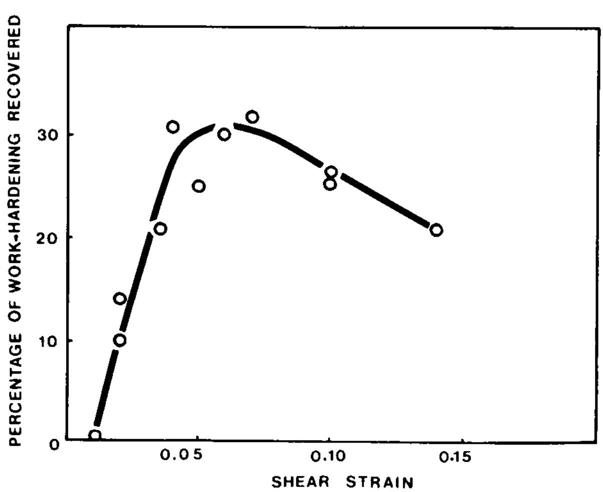

Figure 1.5 - Recovery in Cu-Al_2O_3 Crystals Deformed at 77K and Annealed 3 Hours at 293K [1.42]

Measurements of the activation energy of the recovery process [1.4,25,28] yield values of ∿ 0.9 ev, approximately half that for bulk self diffusion in copper. The recovery rate is particle size dependent, being most rapid for small particles, e.g., BeO and slowest for larger SiO_2 particles. For particles of diameter $\gtrsim 3000$Å, very little recovery occurs [1.25]. It is also found that the recovery depends on the nature of the strain, and a large amount of recovery is found only when single slip occurs in the matrix. Thus, in stage I of Cu-Al_2O_3 crystals, ∿ 40 percent of the work hardening is recoverable, but in stage II, for a similar stress increment, only ∿ 1 percent is recovered [1.18]. Similarly, recovery has not been reported in deformed polycrystalline alloys.

1.3.5 *Temperature Dependence of the Flow Stress of Single Crystals*

The rate of work hardening during the parabolic stage I of dispersion hardened single crystals is found to be very dependent on temperature and strain rate between ∿ 200 - 500K [1.18,27,28], as shown in Figures 1.6 and 1.7. However, strain rate change tests [1.18,29] show that the instantaneous change in flow stress is small, and comparable to pure copper, indicating that the temperature and strain rate sensitivities of the work hardening rate are primarily due to changes in microstructure rather than to changes in dislocation mobility.

1.4 THE MECHANISMS OF PLASTIC DEFORMATION

1.4.1 *Orowan Loop Formation*

When a dispersion hardened alloy is deformed beyond the yield stress, repeated operation of the mechanism of Figure 1.1 leads to the formation of arrays of Orowan loops at the particles. The average number of loops per particle (n_0) is given [1.30] by

$$n_0 = \frac{2r\gamma}{b} \, . \qquad (1.6)$$

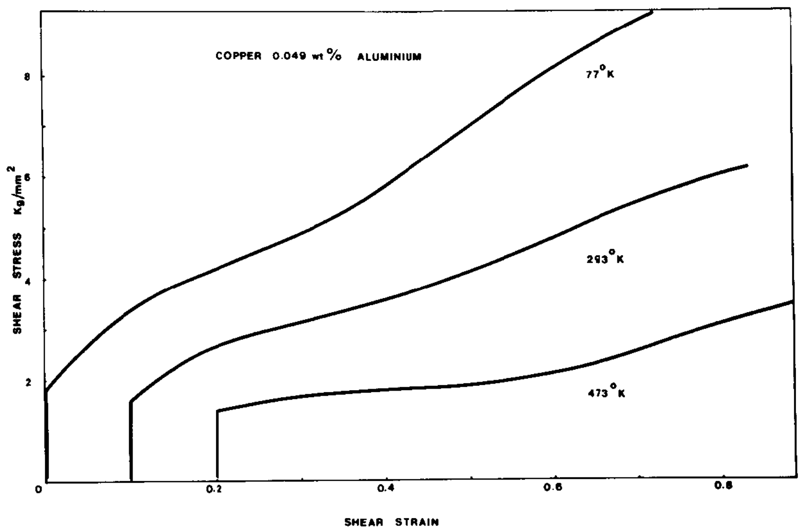

Figure 1.6 - The Effect of Temperature on the Stress-Strain Curves of $Cu\text{-}Al_2O_3$ Crystals [1.18]

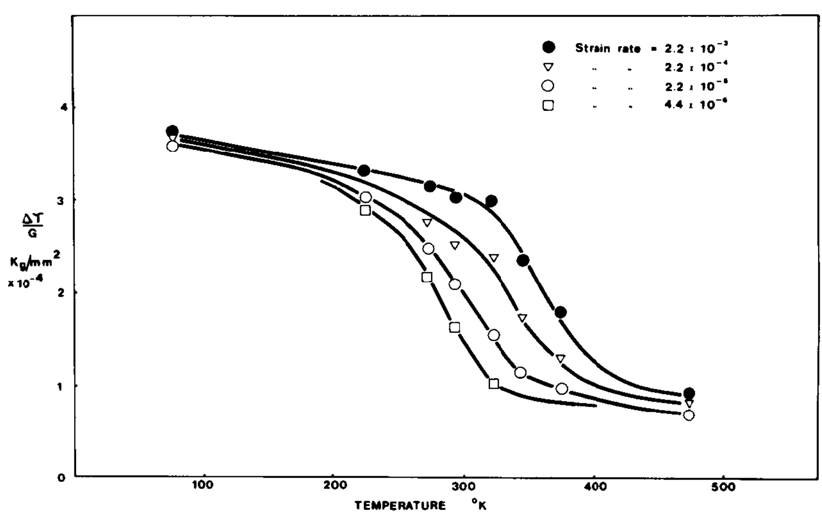

Figure 1.7 - The Work Hardening Increment for $Cu\text{-}Al_2O_3$ Crystals as a Function of Temperature and Strain Rate [1.18]

For particles smaller than the slip line spacing, concentric arrays of Orowan loops are expected (Figure 1.8(a)), whereas for larger particles a more uniform distribution of Orowan loops will form (Figure 1.8(b))

(a) (b)

Figure 1.8 - Orowan Loops at Particles: (a) small particles; (b) large particles

The stresses in and around a particle due to an array of Orowan loops have been considered by several authors. For the uniform distribution of Figure 1.8(b), the local shear stress in the matrix will be [1.30,31]

$$\tau_i \sim \frac{n_0 G b}{2r} , \qquad (1.7)$$

and using (1.5)

$$\tau_i \sim G \quad . \qquad (1.8)$$

Li and Liu [1.32] and Hazzledine and Hirsch [1.33] have considered concentric arrays of loops as in Figure 1.8(a), and the numerical data of the latter authors for the shear stress at the particle interface is shown in Figure 1.9. An analysis of their data shows that to a good approximation, the stress at the interface is given by equation (1.9), the exact value dependent on the applied stress,

$$\tau_i \sim \frac{G b n_0^{3/2}}{2r} , \qquad (1.9)$$

and using (1.6)

$$\tau_i \sim \sqrt{2}\, G \gamma^{3/2} \left(\frac{r}{b}\right)^{1/2} . \qquad (1.10)$$