Electron Microscopy

and Strength of Crystals

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Proceedings of the First Berkeley International Materials Conference "The Impact of Transmission Electron Microscopy on Theories of the Strength of Crystals" held at the University of California, Berkeley, July 5–8, 1961

Preface

The papers and discussions in this book were presented at an international conference held at the University of California in Berkeley on July 5–8, 1961.

The conference was sponsored jointly by the Inorganic Materials Division of the Lawrence Radiation Laboratory, United States Atomic Energy Commission, and the Mineral Technology Department and University Extension, University of California. Mr. Ira Pratt of the Inorganic Materials Laboratory and Mr. William Harris of University Extension were particularly helpful in making the many detailed arrangements.

The purpose of the conference was to bring together all of the significant findings of transmission electron microscopy concerning dislocation and transformation substructures in materials and to discuss the impact of these observations on the theories of the strength of crystals.

In planning and arranging the program for the conference it was decided that both the meeting itself and the book would be more useful if the number of papers was limited. Therefore about twenty people who had been active in the application of transmission electron microscopy to the study of dislocation and transformation substructures were invited to prepare papers on some aspect of the subject. They were encouraged to include significant results of others as well as their own research. In this way it was hoped that no important observations would be overlooked and that the collection of papers would represent a critical review of all the electron transmission observations that had been made up to the time of the meeting.

Whereas the first half of the program was devoted primarily to a review of the experimental work, the second part of the conference dealt primarily with interpretation. Models for yielding, strain hardening, precipitation hardening, and the strength of martensite were reviewed in the light of the electron transmission work.

In the written papers the division between experimental observations and interpretation is less distinct. The first section of the book deals with dislocation substructures and their relation to the understanding of mechanical properties of single phase materials. The second section is devoted to materials in which more than one phase is involved. Discussions which were recorded during the conference and also a few written discussions which were submitted after its close are printed at the end of each paper.

One of the important conclusions that emerged from the conference is the need for extreme care in the interpretation of thin-foil observations. Dislocation arrangements may be changed during thinning and handling of foils and the electron diffraction contrast effects are sometimes too complex for certain correct interpretation. However, there is no doubt that the techniques of making and handling foils and the capabilities of the electron microscopes themselves will continue to improve. The observations described at this conference are certain to stimulate further and more quantitative work.

The editors would like to thank all the authors of papers and those who contributed to the discussions during the conference. It was their excellent cooperation and hard work that made the conference successful.

Berkeley, California

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Part A

Single Phase Materials

Chapter 1

Defects in Quenched and Irradiated Face-Centered Cubic Metals

M. J. WHELAN

I. Introduction

When a metal is quenched from high temperatures or subjected to bombardment by high energy nuclear particles, a large number of point defects in excess of the equilibrium concentration may be trapped in the lattice. In the former case, only vacancies are thought to be produced in appreciable numbers, whereas in the latter case, vacancies and interstitial atoms will be produced in equal numbers in the interior of the metal at collision sites. The changes in physical properties due to the point defects (e.g., electrical resistivity, density, hardness, and plastic properties) and their variation on subsequent annealing have therefore been the subject of much experimental and theoretical investigation in recent years (see the reviews by Cottrell (1), Seitz and Koehler (2), Kinchin and Pease (3), Glen (4), Dienes and Vineyard (5), Cottrell (6)). While much has been learned from such studies concerning activation energies of formation and migration of defects (see references (7–17) for quenching and (18–28) for irradiation effects), the nature and distribution of the defects introduced and the mechanisms of annealing remained speculative in many cases. For example, it was uncertain whether vacancies in quenched metals clustered to form cavities at dislocations (Coulomb and Friedel (29)), dislocation loops (Kuhlmann-Wilsdorf (30)), or jogs at dislocations (Cottrell (6)).

Recently, however, methods of direct observation by transmission electron microscopy have been employed to study a number of problems in quenching and radiation damage. While much remains obscure, many facts of interest have emerged. The object of this paper is to review the electron microscope evidence obtained so far on the microstructure of quenched and irradiated face-centered cubic metals, and to discuss it in the light of evidence obtained from measurement of other physical properties.

II. Defects Produced by Quenching

A. GENERAL

The equilibrium concentration c of vacancies in a metal (the fraction of sites vacant) increases with temperature T according to the equation

$$c = A \exp\left(-E_f/kT\right) \tag{1}$$

where A is an entropy factor (taken to be of order unity), E_f is the energy of formation of a vacancy, and k is Boltzmann's constant. In aluminum, for example, $E_f \simeq 0.76$ ev (Bradshaw and Pearson (9)). Taking $k = 0.86 \times 10^{-4}$ ev degree C^{-1}, we find that $c \sim 10^{-4}$ near the melting point and that $c \sim 10^{-13}$ at room temperature. Assuming that all the vacancies are retained in a quench, the vacancy supersaturation at room temperature ($\sim 10^9$) is very large indeed.

Frank (31) and Seitz (32) first suggested that the excess vacancies might precipitate in the form of discs which would subsequently collapse to form loops of dislocation line, as shown schematically in Figure 1(b–d). This idea was developed by Kuhlmann-Wilsdorf (30), who also pointed out certain differences in the properties of the loops in face-centered cubic metals of different stacking fault energy. Figure 2a shows Thompson's notation (33) for denoting slip planes and Burgers vectors in the face-centered cubic structure. The tetrahedron $ABCD$ has edges parallel to the six close-packed $\langle 110 \rangle$ directions and faces parallel to the four close-packed $\{111\}$ planes. The edge length is equal to the length of the Burgers vector $\frac{1}{2}\langle 110 \rangle$ of a whole dislocation. The midpoints of the faces opposite A, B, C, and D are denoted by α, β, γ, and δ. Imagine that the dislocation loop in Figure 1d lies in the plane BCD of Figure 2a (plane α). If it is formed by removal of part of a B-plane and straight collapse (as shown in Fig. 1d), the Burgers vector of the loop will be $\frac{1}{3}\langle 111 \rangle$ or αA. The dislocation is a Frank sessile dislocation, enclosing a region of intrinsic stacking fault (Frank (34)). However, as pointed out by Kuhlmann-Wilsdorf (30), it may be energetically favorable for such a loop to revert to a whole dislocation line by shear of the C-

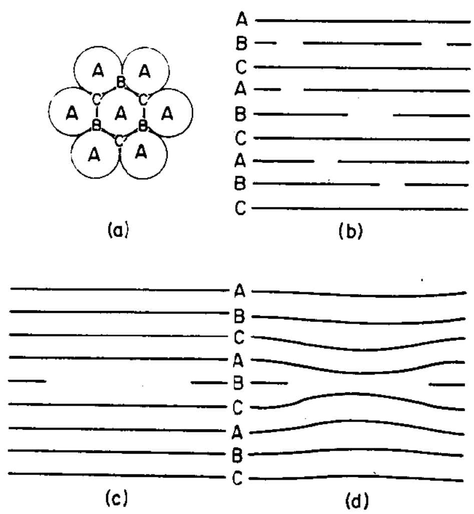

Fig. 1. Schematic diagram illustrating dislocation loop formation by vacancies in a face-centered cubic metal. (a) Stacking sequence (b) random vacancies (c) disc of vacancies (d) collapsed disc with intrinsic stacking fault.

layer bounding the vacancy disc to a B-position, thereby removing the stacking fault. Such a shear is represented by any one of the vectors $B\alpha$, $C\alpha$, $D\alpha$ in Figure 2a, which are Burgers vectors of Shockley partial dislocations glissile in plane α. For example, the reaction $B\alpha + \alpha A \rightarrow BA$ will produce a whole dislocation loop of Burgers vector BA. While the original sessile loop was constrained to lie on α, it will be noted that the whole dislocation loop can glide on the prismatic cylinder containing the loop and the direction of the Burgers vector, and is therefore called a "prismatic" dislocation loop. This type of loop was also referred to as an R-dislocation by Kuhlmann-Wilsdorf (30).