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of twist boundaries were prepared in such a way that initially the boundary region was free from dislocations. This enabled us to study the generation of dense dislocation networks in the course of annealing. The imaging in TEM and the form of the dislocation network will be discussed in some detail for a low-angle (111) twist boundary and a near-coherent twin boundary. For a low-angle (111) twist boundary a hexagonal dislocation network of $a/2 <110>$ screw dislocations is expected to occur. However, TEM imaging of such a network is complicated by the superposition of strain contrast from the dislocations and the interference effects (moire fringes) associated with the twist at the boundary. This may result in different types of images, depending on the diffraction conditions employed.

We found that a hexagonal network can be imaged as a triangular black-white contrast pattern, as if the hexagonal network is dissociated into a triangular network of partial screw dislocations. For a near-coherent twin boundary a dislocation network consists of $a/6 <112>$ screw dislocations. In this case a step in the boundary is associated with each dislocation. It can be shown that in order to preserve the average boundary normal, a network in the form of a regular six-star pattern should occur, as observed. Also in this case the network image depends on the diffraction conditions employed.

Finally some observations on low-angle phase boundaries in (111) Au/AuPd bicrystals will be discussed. Here the boundary consists of a rotation as well as a dilatation component, and it is found that the latter component disappears first on annealing the bicrystal.


THE USE OF TEM IN THE STUDY OF ORDERING ALLOYS

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At high temperature Cu$_2$NiZn possesses an f.c.c. disordered structure. Below the first critical temperature, $T_{c1}=774K$, a modified L1$_0$ structure exists, in which the Zn atoms occupy one of the four interpenetrating simple cubic sublattices, while the Cu and Ni atoms are still randomly distributed over the remaining three sublattices. Below a second critical temperature, $T_{c2}=598K$, a modified L1$_0$ structure exists, in which the Cu and Ni atoms each occupy their own sublattice.

Deformation of f.c.c. metals and disordered solid solutions occurs by the movement of unit dislocations, consisting of two Shockley partials, separated from each other by a stacking fault. Slip of such a unit dislocation in an ordered alloy would create a sheet of antiphase boundary (APB) behind it. The disorder associated with this APB can be restored by slip of a second unit dislocation, with the second dislocation assumed to be the first dislocation. Such a dislocation pair, consisting of two unit dislocations separated by an APB is called a superlattice dislocation. Each of these unit dislocations is split up into two Shockley partials, separated by an APB and a stacking fault. Anisotropic elasticity theory provides the relation between the separation of the Shockley partials and the APB energy, stacking fault energy and the character of the dislocation. Several superlattice dislocations in Cu$_2$NiZn have been characterized and imaged in the weak beam mode. The APB energy of dislocations has been determined from the electron micrographs by means of a densitometer. A corresponding APB energy of $89 \pm 9 mJ/m^2$ has been calculated. However the splitting of each unit dislocation into two Shockley partials could not be observed, which means that their separation is less than about 0.5 nm indicating a high stacking fault energy compared to Cu (40 mJ/m$^2$).

An extended dislocation node is the product of a reaction between two unit dislocations having different Burgers vectors and slipping on different type (111) planes. Several theoretical treatments of the node problem have been published in the literature, resulting in a similar prediction for an extended dislocation node of screw character. The predicted dimensions for edge character nodes differ a great deal, but these nodes are only seldom observed. About 17 extended dislocation nodes have been characterized in disordered Cu$_2$NiZn, from which it can be concluded that the models of Siems, Jøssang et al. and Scattergood et al. provide good agreement, whereas the models of Brown et al. and Milon et al. exhibit intrinsic discrepancy. A stacking fault energy of $32 \pm 7 mJ/m^2$ is obtained for Cu$_2$NiZn.

The fault energies of Cu$_2$NiZn as determined by TEM are of importance, as will be explained with two examples. Firstly, the value of the APB energy is calculated using effective interatomic potentials of the transition metals which are computed within the framework.
of the pseudopotential approximation based on a transition-metal-model-poten-
tial of the Heine-Abarenkov-Shaw type. The experimentally determined APB energy of 89 mJ/m² agrees very well with the ab initio calculated value of 94 mJ/m². Secondly, with these interatomic potentials long-range and short-range order parameters of Cu₂NiZn have been calculated as a function of temperature, using the Cluster Variation Method. The strengthening theories of Moine et al. and Rudman for ordering alloys have been used to predict the yield strength of Cu₂NiZn as a function of the quench temperature, using the experimentally determined fault energies and the calculated order parameters. The agreement with the experimentally determined yield strength of the alloy is reasonable.

Measurements of physical properties (resistivity, lattice parameter, micro-
hardness, etc.) result in a determination of the critical temperature for ordering, but they cannot reveal the extent of a two-phase region, which is present in the case of a first-order transition. TEM, however, can supply direct evidence of a two-phase region by imaging the thermal APB domain with three independent super-
lattice reflections. A modification of this method is applied to obtain the quasi-binary cross-section Cu₅₀Ni₅₀₋ₓZnₓ. The upper boundary of this two-phase region as determined by TEM agrees well with resistivity measure-
ments.


FIT AND MISFIT IN IMAGING OF "UNFIT" CRYSTALS

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After a short review of the various types of two-dimensional defects, the imaging properties are summarized as derived for two-beam conditions from dynamical diffraction theory including anomalous absorption.

Defects to be discussed are antiphase boundaries stacking faults, domain- and twin boundaries, inversion boundaries and grain boundaries. It will be shown how these are usually imaged as fringe patterns, and how the image characteristics can be utilized to derive crystallographic characteristics of the defects. Three main types of patterns can be dis-
tinguished: α-type fringes, δ-type fringes, and thickness or wedge fringe patterns.

The availability of high resolution microscopes now also enables these defects to be observed in the structure projection imaging mode. The bright- and dark-field medium magnification images can now be suitably complemented with multiple beam imaging and magnifications of million times. In speci-
mens of appropriate orientation, thickness and structure type the above mentioned defects are directly revealed. The ef-
flect of objective aperture size and specimen thickness will be illustrated on the imaging of structures and struc-
tural defects such as antiphase boundaries, etc. in ordered alloys. Gold manga-

nese and γ-brass will be used as exam-


THE VOLTAGE DILEMMA: APPLICATION TO MATERIALS SCIENCE

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The first electron microscope was built in 1931 by E. Ruska and M. Knoll. At that time they were delighted to dem-