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THE STRUCTURE OF TWINS IN FE–NI MARTENSITE

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Abstract—This paper corrects an error in a previous paper by Déchamps and Brown and presents a much improved model for the structure of twins in Fe–Ni martensite. The model is based upon the dissociation of screw dislocations under the stress due to the martensitic transformation.

Résumé—Dans cet article, nous corrigeons une erreur d'un article précédent de Déchamps et Brown et nous présentons un modèle amélioré de la structure des macles dans la martensite Fe–Ni. Ce modèle repose sur la dissociation des dislocations vis sous l'effet de la contrainte produite par la transformation martensitique.

Zusammenfassung—Mit dieser Arbeit wird ein Fehler in einer früheren Arbeit von Déchamps und Brown korrigiert. Ein weitgehend verbessertes Modell wird für die Struktur von Zwillingen im Fe–Ni-Martensit vorgeschlagen. Das Modell geht aus von der Aufspaltung von Schraubenversetzungen unter der von der martensitischen Umwandlung erzeugten Spannung.

INTRODUCTION

In a previous paper by two of us [1], the structure of twins in Fe–Ni martensites was described as a three-dimensional stack of partial dislocations with Burgers vectors $b = a/6\langle 111\rangle_{\text{b.c.c.}}$. They are supposed to be arranged so that the crystallographic twinning plane (i.e. the plane in which the stacking faults are extended) is different from the composition plane, although it belongs to the same $\langle 111\rangle_{\text{b.c.c.}}$ zone. This structure was based on two results: [1] the correlation of the mirror plane in electron diffraction patterns with a well-defined edge of the twin; Déchamps and Brown assumed that the mirror plane coincides with the crystallographic twinning plane and it is found that this plane always contains the direction of the edge of the twin. [2] The second result derived from a trace analysis which showed that the edge of the twin is a $\langle 111\rangle_{\text{b.c.c.}}$ direction and that the composition plane is a $(11\bar{2})_{\text{b.c.c.}}$ plane which does *not* coincide with the mirror plane.

There is an error in this line of argument, namely in the assumption that the mirror plane in the electron diffraction pattern coincides with the crystallographic twinning plane, i.e. the plane on which the shear displacement occurs. Consider the crystal and its twin. They have a common threefold axis of symmetry, which lies in the mirror plane and in the b.c.c. case coincides with the twinning direction. It follows that in reciprocal space, there are three equivalent mirror planes and observation of one of them in a diffraction pattern does not determine which of the three has acted as the plane on which the twinning shear has physically occurred. Thus the error in the analysis of Déchamps and Brown lies in the assumption that observation of the diffraction pattern alone determines the plane on which the twinning shear occurs. The twinning shear can in principle be established only by observation of surface displacements, for example by replicas taken from pre-polished and transformed specimens: it cannot be determined by electron diffraction (or by direct lattice resolution in real space!).

It follows that the conclusion of Déchamps and Brown must be altered. The twins are still to be regarded as accumulations of screw dislocations, but the plane on which they accumulate can now coincide with the plane on which the twinning shear takes place; this of course provides good elastic accommodation. The new proposal is shown in Fig. 1. In Fig. 1(a) are shown two screw dislocations of opposite sign, each one split in one of two ways. The applied stress produces a force on each as shown by the arrows. Figure 1(b) shows one possible consequence of the passing of the screw dislocations of Fig. 1(a): partial annihilation results in a stepped stacking fault which is a single-layer twin. Figure 1(c) shows the possible configuration of faults when several screw dislocations have interacted and split; a thin, imperfect twin is formed with striations parallel to the twinning direction. The composition plane is parallel to the plane on which the twinning shear occurs, so the strain is well accommodated. The condition for the formation of such twins is simply that the stacking-fault energy should be low enough and the stress high enough for the fault in Fig. 1(b) to be extended; if we assume that the stress acting is determined by the transformation strain (ϵ) of the martensite, then dislocated martensite will be replaced by twinned marten-

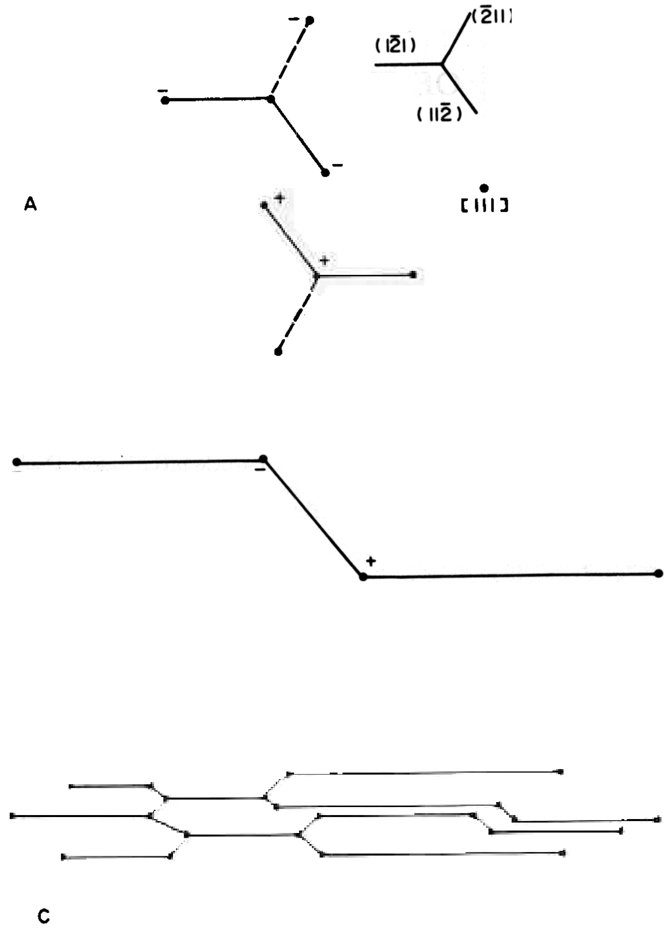

Fig. 1. (a) Dissociated screw dislocations in b.c.c. Dashed lines represent alternative splittings. Normal to paper [111], planes of stacking fault, $\{\bar{2}11\}$. (b) Screw dislocations interact to form extended stepped fault. Partial dislocations labelled + and − cancel. (c) Irregular thin twin formed by accumulation of screw dislocations.

site when the stacking-fault energy, γ, is lower than some value. A simple estimate for the critical value is $\gamma \lesssim \epsilon\, Gb$, which is equal to about $100\,\mathrm{mJm^{-2}}$, a not unreasonable value to judge from the figures quoted in the literature. (In this calculation, ϵ is taken to be that of the shape deformation in a martensite plate with an aspect ratio of about 1/10).

Thus the revised model explains most features of the observations in a satisfactory way; in particular it explains how the twins can thicken by the accumulation of screw dislocations. It is clear that the twinning observed here (often called 'ribbon twinning') is quite different in origin from the twins in classical twinned martensite. The latter occupy a definite volume fraction and extend fully across the martensite plate; they are also very regularly arranged. The ribbon twins are most easily understood as accommodation effects by plastic flow following the formation of the martensite plate.

REFERENCE

1. M. Déchamps and L. M. Brown, *Acta metall.* **27**, 1281 (1979).