A DIRECT ANALYSIS OF TWINNING IN A LOW ALLOY MARTENSITE

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INTRODUCTION

The deformation processes involved in the formation of low alloy martensites appear from microstructural observations to be essentially slip. However, extremely high dislocation densities are generally observed, such that the arrangement of the dislocations is very confused and does not give useful information on the slip systems involved. The fact that the dominant deformation process in low alloy martensite is slip is not surprising since the Ms temperatures concerned are relatively high when compared with fully twinned plate martensites (1). Despite the high Ms, however, incipient twinning can often be observed in low alloy martensites (2, 3). These twins sometimes extend only partially across the martensite units (3) and the inter-twin spacings are generally larger and more variable than those intuitively expected from observations on well-defined plate martensites. It is not apparent whether these twins represent a mode of lattice invariant shear (and hence whether they can be useful in the application of the phenomenological theory of martensite formation) or whether they are simply manifestations of accommodation effects.

Unfortunately, to be able to differentiate between transformation and accommodation twinning, detailed crystallographic analysis between the parent and product phases is essential. The lack of adequate quantities of the parent austenite phase and the scale of the martensite microstructure in low alloy steels has excluded such an analysis in the past. In the present work, a novel method is used to enable the retention of experimentally expedient quantities of austenite, thus enabling a direct analysis of the twins in a low alloy martensite.

EXPERIMENTAL METHOD AND TECHNIQUES

An experimental Fe-0.43C-3.0Mn-2.02Si steel of known transformation characteristics (4) was used in the present study. The Ms temperature of this alloy is 220°C and isothermal transformation at 350°C results in the formation of upper bainite. The upper bainite in this steel is exceptional in the sense that the high silicon content prevents the formation of carbides. Thus, the carbon which is partitioned into the residual austenite upon the formation of bainitic ferrite stabilises the austenite and causes its retention following quenching to room temperature. This characteristic was exploited as described below.

A specimen was austenitised in a dynamic protective argon atmosphere at 1100°C for 20 mins followed by a direct quench into oil at 140°C. After holding at 140°C for 40 s, it was immediately up-quenched into a tin bath held at 350°C for isothermal transformation of the remaining austenite to upper bainite. After holding at 350°C for 25 mins, the specimen was finally water quenched. The direct quench to 140°C resulted in partial transformation to martensite, to an estimated extent of 50% by volume. This is because 140°C is below the Ms but well above the Mf temperature of this alloy. The specimen was first transformed to martensite so that the carbon content of the martensite was the same as that of the alloy. The subsequent up-quench resulted in the transformation of the remaining austenite to upper bainite, with the accompanying carbon enrichment of the austenite between the martensite units, and between the bainite plates. Thus, when the specimen was finally quenched to room temperature, the inter-martensite residual austenite was fully retained. Under normal circumstances this would have transformed to martensite.

Specimens for examination by transmission electron microscopy were prepared in the manner described in ref. 4. A Philips EM300 transmission electron microscope operated at 100kV was used for this work.

RESULTS AND DISCUSSION

The orientation relationship between the martensite and austenite was found by simple electron diffraction to be that of Kurdjumov and Sachs (5), (K-S), i.e.

$$(111)_\gamma \;//\; (0\bar{1}1)_{\alpha'}$$

$$[\bar{1}01]_\gamma \;//\; [\bar{1}\bar{1}1]_{\alpha'}$$

The above variant will be used as the standard variant in the analysis that follows. While it is noted that simple electron diffraction of this kind cannot give accurate orientation relationships, it will become clear later that the orientation has to be K-S. This conclusion comes from the observation of twin-related martensite variants a situation which has been shown to arise (6) only with the K-S orientation relationship.

Owing to the high silicon content of the alloy used, the formation of martensite is not accompanied by autotempering (4). However, due to the up-quench component of the heat treatment used for the present study, the martensite was found to be tempered, as shown in figs. 1a and 2a.

Single surface trace analysis was carried out to establish the twin plane and its relation to the austenite lattice. The analysis was always in terms of the standard correspondence stated above. The great circle representing the locus of the twin plane pole was plotted consistently and unambiguously with respect to the austenite martensite and martensite twin lattices. The results are presented in figs. 1 and 2.

For martensite twins to be transformation twins, certain symmetry requirements have to be satisfied (7). Assuming that the twins are of type 1 (involving a rotation of 180° about the pole of the twin plane), the twins have to relate adjacent martensite regions whose c-axes are variants of the c-axis of the Bain distortion. This means that the twin plane has to be a plane of mirror symmetry with respect to the austenite lattice after transforming through the appropriate correspondence matrix. Furthermore, for the twinning system to be considered as an intrinsic transformation inhomogeneity, the twin plane must, in all cases, correspond to the same austenite mirror plane when the analysis is in terms of a standard correspondence. Thus specific Miller indices can be assigned to the twin plane (for the standard variant) and these indices have to be the same (not just of the same form) for every case examined if the twins are to be considered as transformation inhomogeneities. Substitution into the phenomenological theory easily demonstrates that lattice invariant shears on different planes of the same form give different (crystallographically non-equivalent) solutions.

In figures 1 and 2 the evidence from the diffraction patterns is plotted on stereograms which also have the poles of the austenite and martensite matrix plotted in the standard K-S variant. The great circles (A) and (B) represent the observed austenite and martensite matrix zones respectively, (D) represents the possible locus of the twin plane normal and (C) refers to the zone of coincidence. The zone of coincidence is defined as the zone obtained by superimposing the martensite matrix and twin stereograms in the correct relative orientation such that on this zone, twin and matrix poles of the same form are coincident. Furthermore, this zone must also contain all the coincident $\{1\bar{1}0\}_\alpha$ and $\{112\}_\alpha$ poles. Assuming that the twin plane will either be $\{1\bar{1}0\}_\alpha$ or $\{112\}_\alpha$, and that the twin boundary corresponds to the twinning plane, the intersection of (C) and (D) defines the twin plane. The former assumption takes account of the fact that the $\{1\bar{1}0\}_\alpha$ and $\{112\}_\alpha$ planes are the most likely twin planes in BCC lattices (8) while the latter assumption is based on the fact that the energy of a twin boundary is a minimum when the composition plane coincides with the twin plane (9).

In fig. 1 the twin plane is seen to be $(\bar{1}12)_{\alpha'} = (0\bar{1}1)_\gamma$ while that in fig. 2 is $(112)_{\alpha'} = (\bar{1}01)_\gamma$. This is despite the fact that they have both been plotted in the standard variant of the orientation relationship. In fig. 1 the twin is not a variant of K-S while in fig. 2 it is. We can therefore conclude that since the twin plane does not uniquely correspond to a particular austenite mirror plane, the twins cannot be an intrinsic transformation feature and must be attributed to accommodation effects.

Some rather spectacular physical evidence for this can be seen from the micrographs of figs. 1 and 2. In fig. 1 the martensite matrices of the adjacent martensite

plates are not twin related. The incidence of twinning is observed to be low and the twins are fine. On the other hand, in fig. 2, the adjacent martensite plates are twin related and dark field imaging shows extensive twinning; often large areas of the plates are twinned. In the latter case, since the two martensite variants are twin related, the nucleation of mechanical twins would be easier, and furthermore, the transmission of deformation across twin related lattices would also be easier. Hence accommodation of transformation strains would be easier for twin related martensite variants.

SUMMARY

It has been found that the twinning observed in the martensite of an Fe-Mn-Si-C low-alloy steel is an accommodation phenomenon since the twin planes show inconsistent correspondence relative to austenite when analysed in terms of the standard variant of the orientation relationship.

The extent of twinning can be understood on the basis of the above conclusion and upon considering the inter-martensite orientations.

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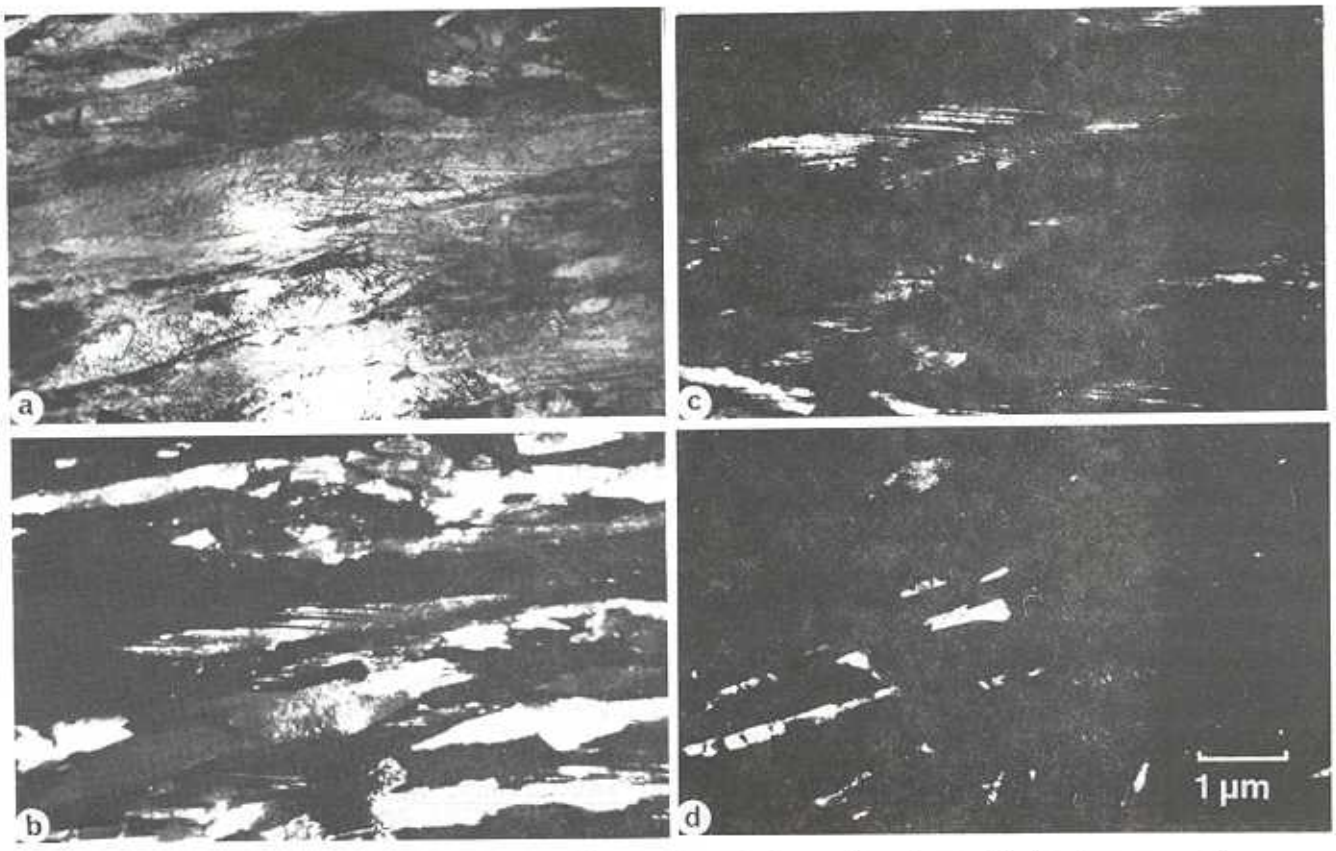

Figure 1 a) Bright Field Image; b) Matrix Martensite dark field image; c) Martensite twin dark field image; d) Retained Austenite dark field image.

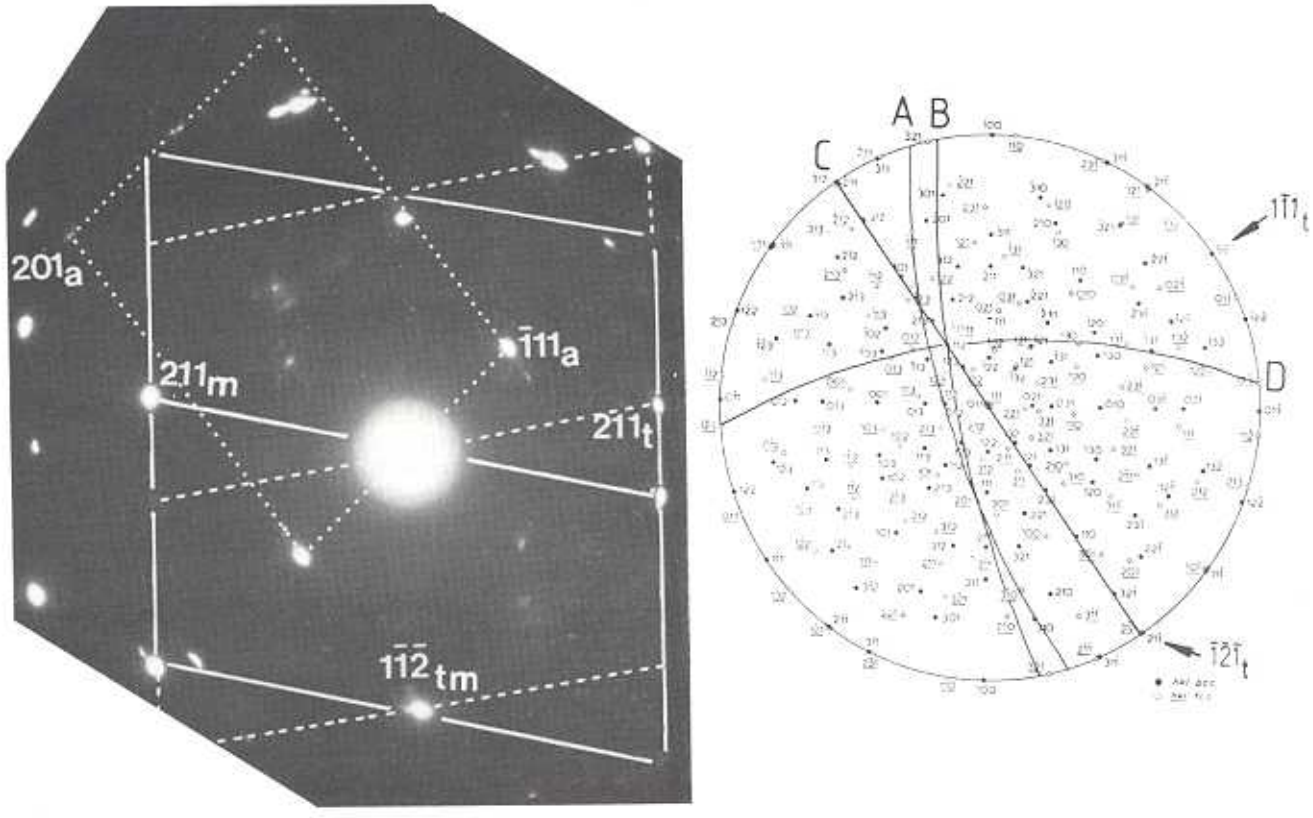

Figure 1 e) Diffraction pattern showing twin-related <135>α zones and <123>γ zone and corresponding stereographic analysis.

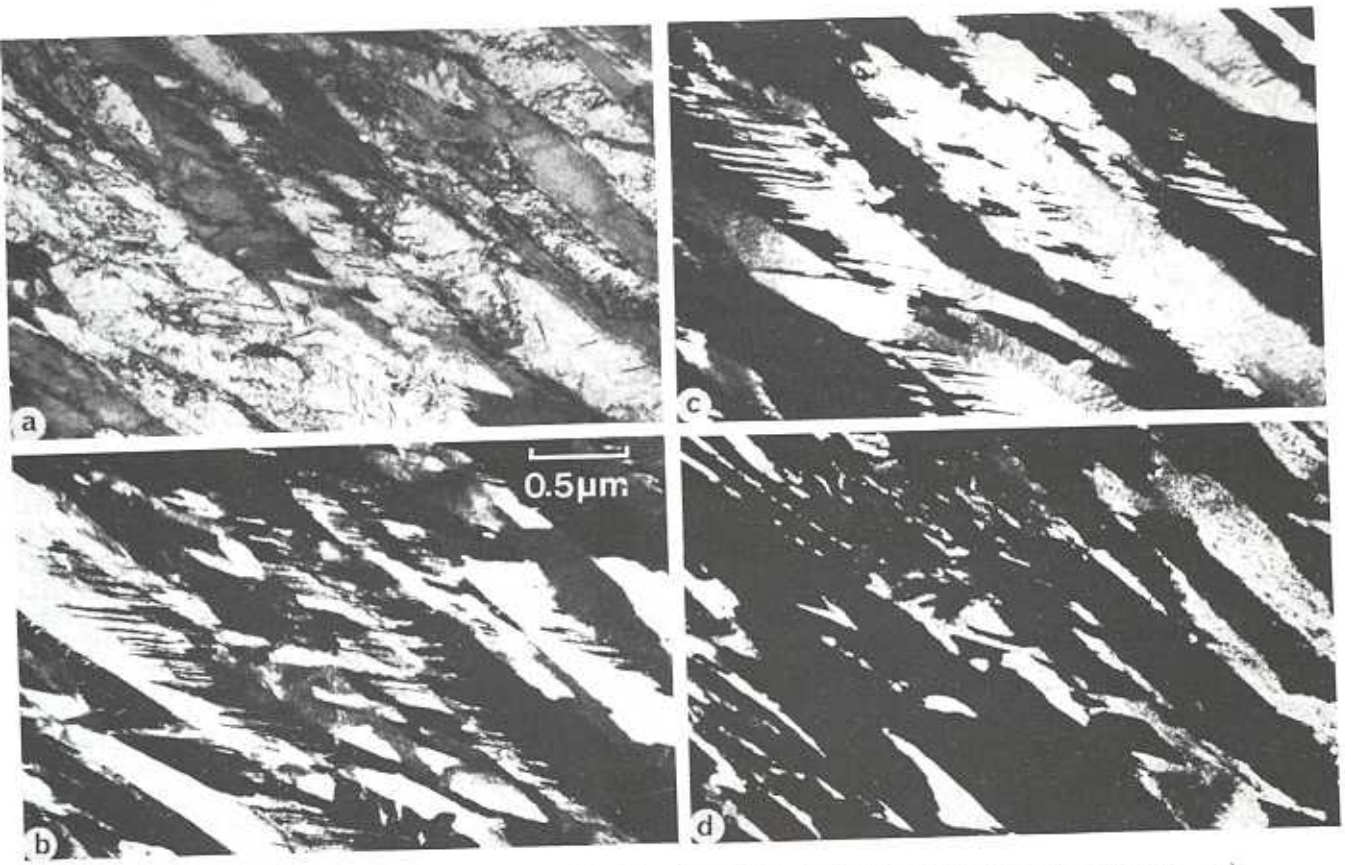

Figure 2 a) Bright Field Image; b) Martensite matrix dark field image; c) Martensite twin dark field image; d) Retained Austenite dark field image.

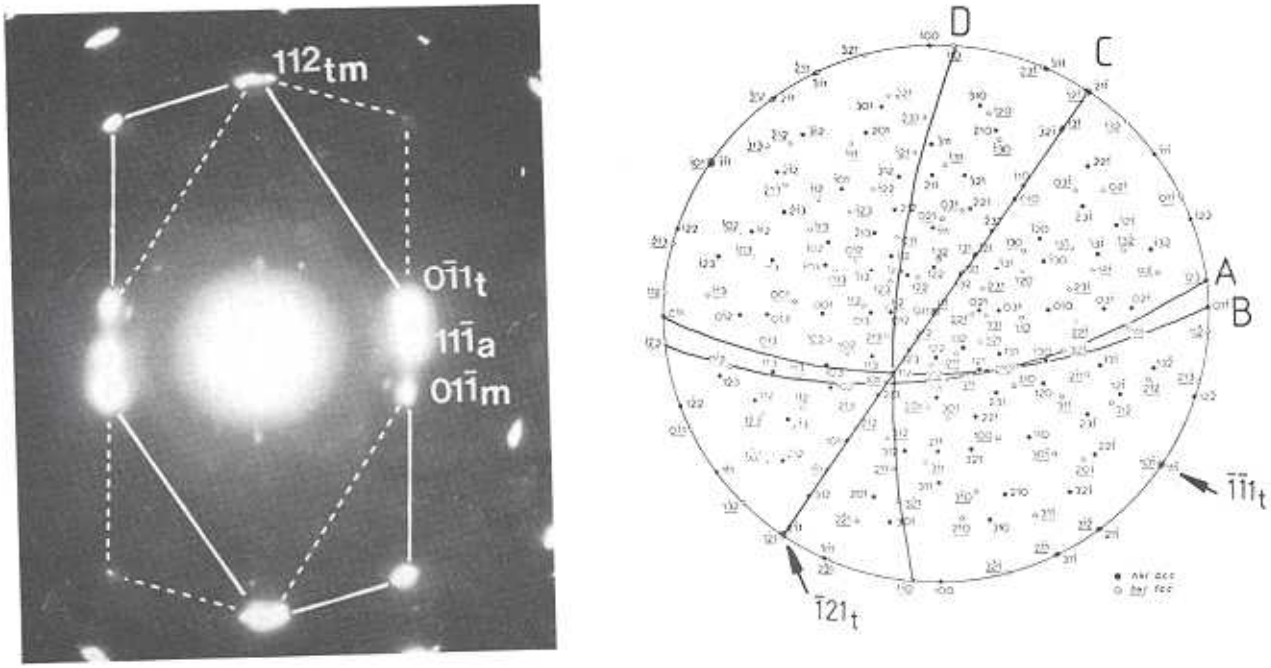

Figure 2 e) Diffraction pattern showing twin-related <113>α' zones and corresponding stereographic analysis.

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