NEW CRYSTALLOGRAPHIC AND MORPHOLOGICAL OBSERVATIONS ON THE MARTENSITE TRANSFORMATION IN AISI 4340 TYPE STEEL

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INTRODUCTION

AISI 4340 steel is a high-strength medium-carbon low-alloy steel used extensively in the quenched and tempered condition. Its composition places it in the interesting transition region between the lath and plate morphological variants of martensite. This region has not been investigated in detail; generally, as in the case of AISI 4340, the microstructure has been interpreted in terms of mixtures of lath and plate martensites (1).

However, the existence of lath martensite as a distinctly separate morphology has recently been questioned (2) in view of the fact that inter-lath retained austenite films have been detected in several directly quenched low-alloy steels (3). Classically, such martensite refers to parallel formations of lath-shaped units with the aggregate of parallel laths being termed the lath packet. With this type of martensite it is not entirely clear whether the individual laths constitute the fundamental transformation unit or whether the packet as a whole represents the transformation unit, with the individual laths being some form of inhomogeneous sub-structure (4). In the former case the lath boundaries would constitute the habit plane, whereas in the latter case these boundaries would represent a polygonised dislocation arrangement linked with transformation strain.

The present work examines the microstructure in the transition region between the plate and lath martensites using transmission electron microscopy.

EXPERIMENTAL

The alloy composition in the electro-slag refined condition was:

C	Mn	Si	Ni	Cr	Mo	
0.38	0.59	0.37	2.89	0.84	0.60	wt. pct.

For the purposes of microstructural examination the alloy was hot worked to 3mm diameter rod. Austenitising treatments were carried out with the specimens sealed under a partial pressure of Argon in quartz tubes. These tubes were fractured before quenching the specimens into water.

Thin foil specimens for transmission electron microscopy were prepared from 0.25mm discs slit from the heat treated 3mm rod under conditions of flood lubrication. The discs were subsequently thinned and electro-polished in a twin-jet polishing unit using a 25% glycerol, 5% perchloric acid and 70% ethanol mixture at room temperature and 55V. The foils were examined in a Philips EM300 transmission electron microscope operated at 100kV.

RESULTS AND DISCUSSION

Since the microstructure is a mixture of lath and plate martensites, it was possible to examine both morphologies in their partially developed states.

Fig. 1a illustrates the typical lath martensite microstructure observed, with a parallel formation of grain boundary nucleated dislocated laths exhibiting approximately the same crystallographic orientation in space. The small inter-lath misorientations are more apparent in the corresponding dark-field image, fig. 1b. It should be noted that the lath widths are of the order of 0.25μm, as epxected for such martensites (5). Fig. 1c shows a higher magnification image of the packet transformation front normal to the major growth direction. It was observed that the front terminated in a distinctly jagged profile, with the lath boundaries coinciding exactly with the positions of the sharp troughs. Retained austenite films, which could be imaged at packet and plate boundaries (fig. 1d) could not be found within the lath packet.

The absence of inter-lath retained austenite, together with the above morphological observations, suggest that the lath packet is the fundamental transformation unit. If the laths themselves represented the basic martensite unit, the results would have to be interpreted in terms of co-operative growth of adjacently nucleated laths (in the same crystallographic orientation) although the driving force for such an eventis unclear since the transformation displacements would be additive rather than mutually compensating, as in the case of certain twin-realted low-alloy martensites (6). If, on the other hand, the packet delineates the fundamental transformation unit, the jagged transformation front can be visualised as a degenerate 'plate' tip - the reasons for such a degeneracy will be discussed later. The lath boundaries could then be attributed to ordered accommodation effects occurring at the troughs in the transformation front. The nature of such effects has yet to be resolved but the fact that they are only linked with the troughs and not with the equally sharp peaks can be understood since the advancing peaks are mainly surrounded by relatively soft austenite. The latter would enable the relatively easier dissipation of accommodation strains compared with the case of the troughs which are essentially in an environment of martensite, due to their complementary geometry.

The above arguments were further validated when the tips of lenticular plates were examined, fig. 2. It was found that these also had a jagged profile, exactly similar to that observed for the lath packets. Furthermore, low angle boundaries could be observed trailing from the troughs (arrowed in fig. 2b) although these were not as clearly defined as in the lath packet of fig. 1. In many instances the habit plane traces of the macroscopic plates and lath packet plane traces were found to be parallel within a given prior austenite grain, suggesting that the packet and plate morphologies may be crystallographically identical, although confirmation of this would require a more detailed analysis.

It was found that at small austenite grain sizes (100μm following austenitisation at 1100°C @ 5 mins) many of the plates were grain boundary nucleated and exhibited a half-lens morphology, fig. 2a. On the other hand, when the specimens were austenitised at 1200°C for 130 mins to give an austenite grain size of 1100μm, most of the plates were intra-granularly nucleated and exhibited a generally lenticular morphology with both tips having a jagged formation, fig. 2c.

It is considered that the jagged appearance of the transformation fronts arises due to the energetically unfavourable situation represented by the high interface curvature (and associated departure of the interface plane from the habit plane) at a monotonically tipped plate. The situation would deteriorate as the aspect ratio of the plate decreased. It is possible that an overall reduction in the amount of relatively incoherent interface would follow if the singular tip degenerated into a series of smaller tips to give a jagged front and reduce the amount of local interface curvature. This phenomenon is somewhat analogous to the situation that arises in deformation twinning where the lens shape gives rise to higher energy (relative to the coherent twin boundary) incoherent twin boundary area.

The lath and plate morphologies observed in the present work can be rationalised as follows:

a) The lath packet would be nucleated at the prior austenite grain boundaries at temperatures high in the martensite transformation range (Ms-Mf). If the packet is then considered as a plate of low aspect ratio, the results would be consistent with the high formation temperature which would mitigate the strain energy control of morphology. The high temperature of formation of the 'lath packet' would also be consistent with the fact that lath martensites are generally observed in steels with high Ms temperatures.

b) As the temperature decreases within the Ms-Mf range, the strain energy control of morphology would become more stringent and the aspect ratio of the martensite plate is expected to increase. Thus we observe the distinctly lenticular morphologies of the grain boundary nucleated plates, an effect which is even more obvious with the intra-granularly formed plates which probably form at the lowest of transformation temperatures.

Occasional twinning was observed in some of the plates, fig. 3. Such twinning extended only partially across the plates and involved a single variant of the twin plane. It was observed that in the majority of cases the twins were associated with adjacent twin related martensite plates. This can best be understood by considering the twins as accommodation defects, since deformation by twinning would be easiest when adjacent martensite units are twin related - this would remove the necessity to nucleate twins. It was also found that the 'lath packet' did not in general show twinning, a fact which is consistent with their higher formation temperature. The latter would favour slip relative to twinning deformation and the probability of impingement would also be lower due to the low degree of transformation.

CONCLUSIONS

The morphology of the martensite in AISI 4340 steel has been rationalised and it appears that the lath packet is in fact a fundamental transformation unit rather than individual laths so that the former can be considered as a low aspect ratio plate formed at the highest of martensite transformation temperatures. An observed degeneracy in the plate tip is thought to be attributable to the need to minimise the high interface curvature associated with a singularly tipped plate.

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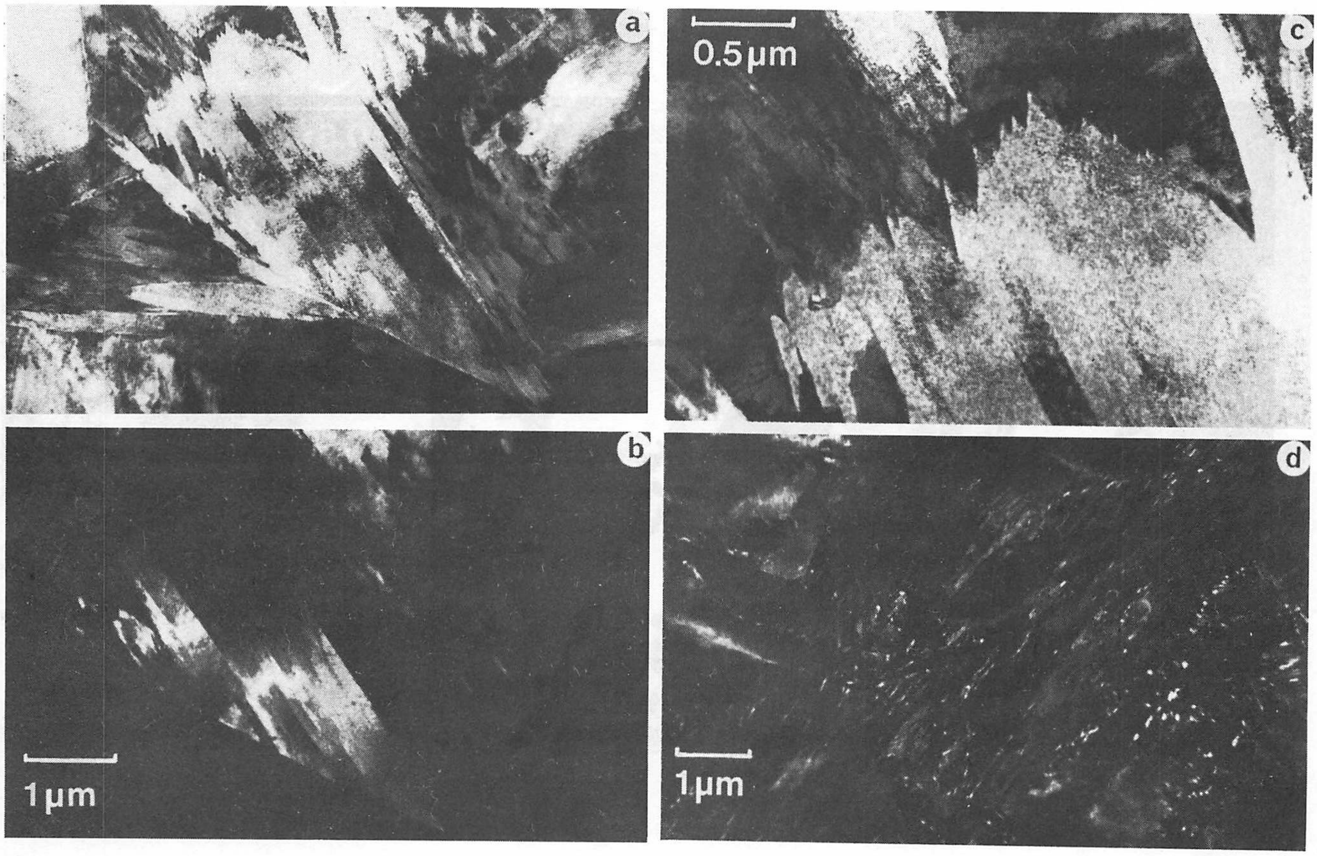

Figure 1 a) Lath Packet, Bright Field Image; b) Corresponding dark field image; c) Higher Magnification image of transformation front; d) Retained Austenite dark field image.

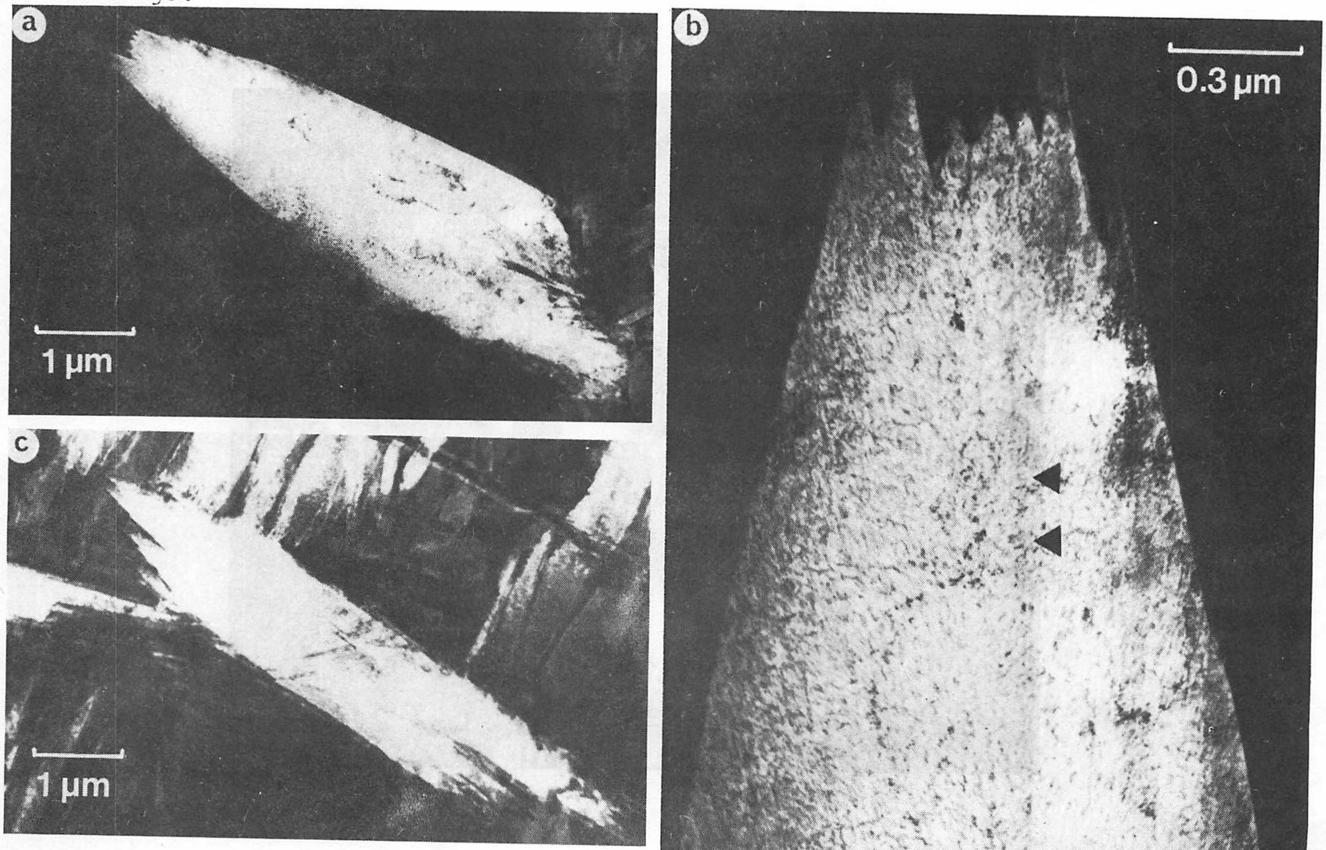

Figure 2 a) Lenticular Plate, Bright field image; b) Higher Magnification image of plate tip; c) Intragranular platelet, bright field image.

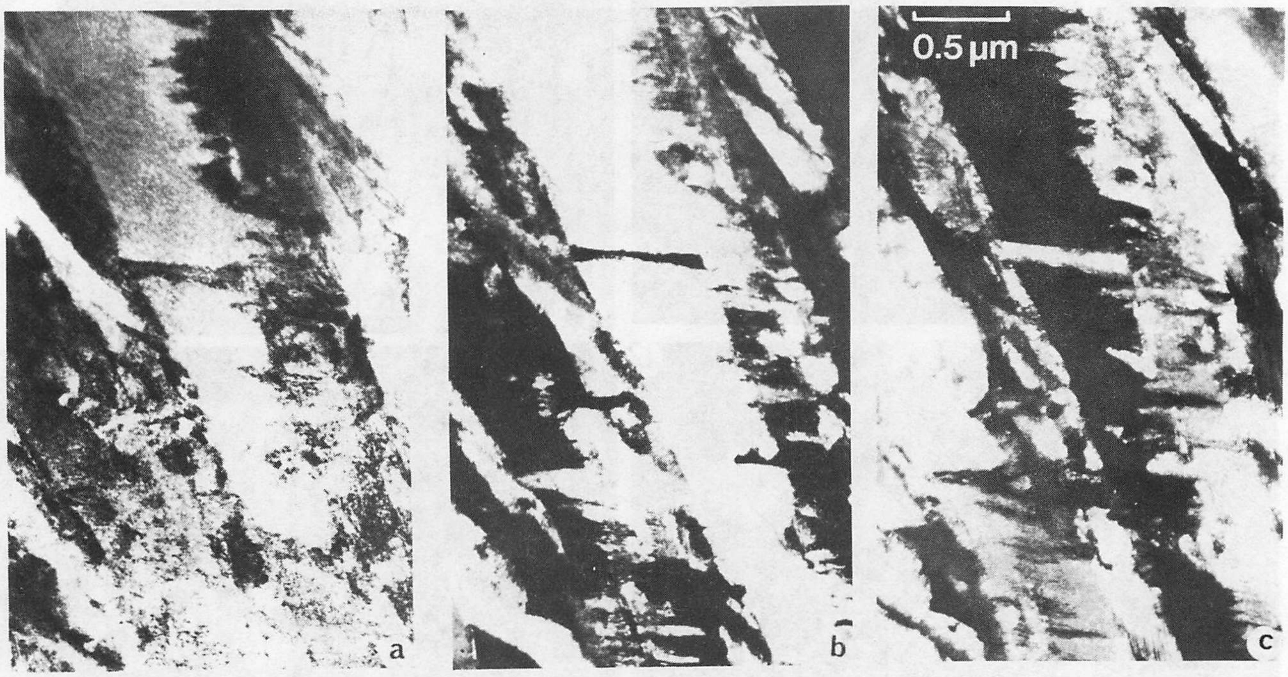

Figure 3 a) Bright Field Image; b) Martensite Matrix dark field image; c) Martensite twin dark field image.

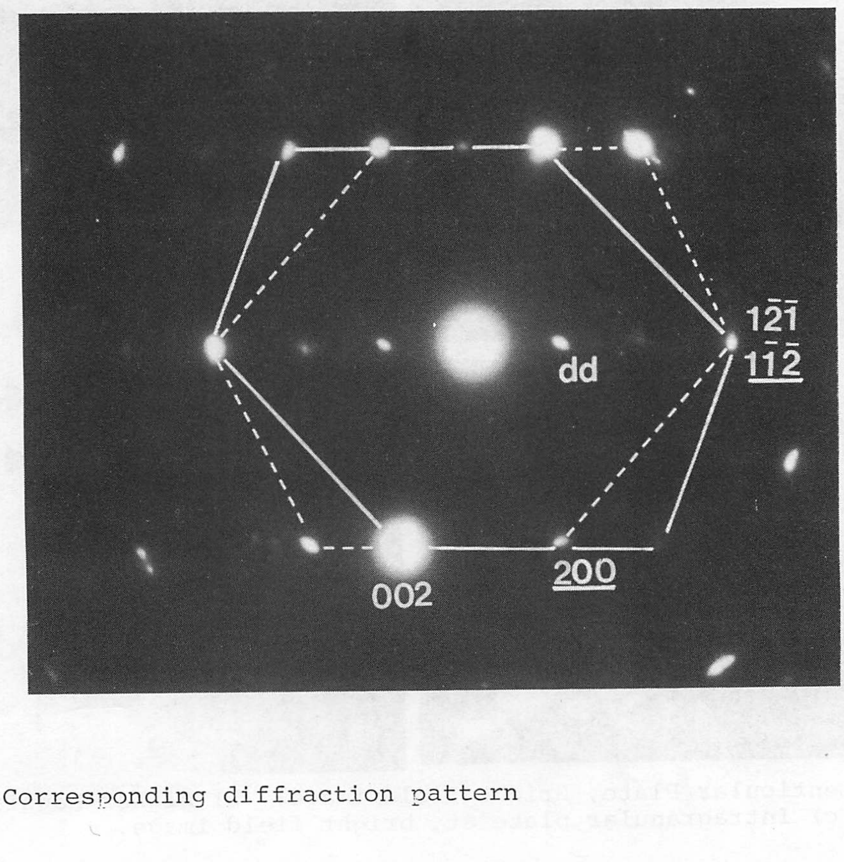

Figure 3 d) Corresponding diffraction pattern