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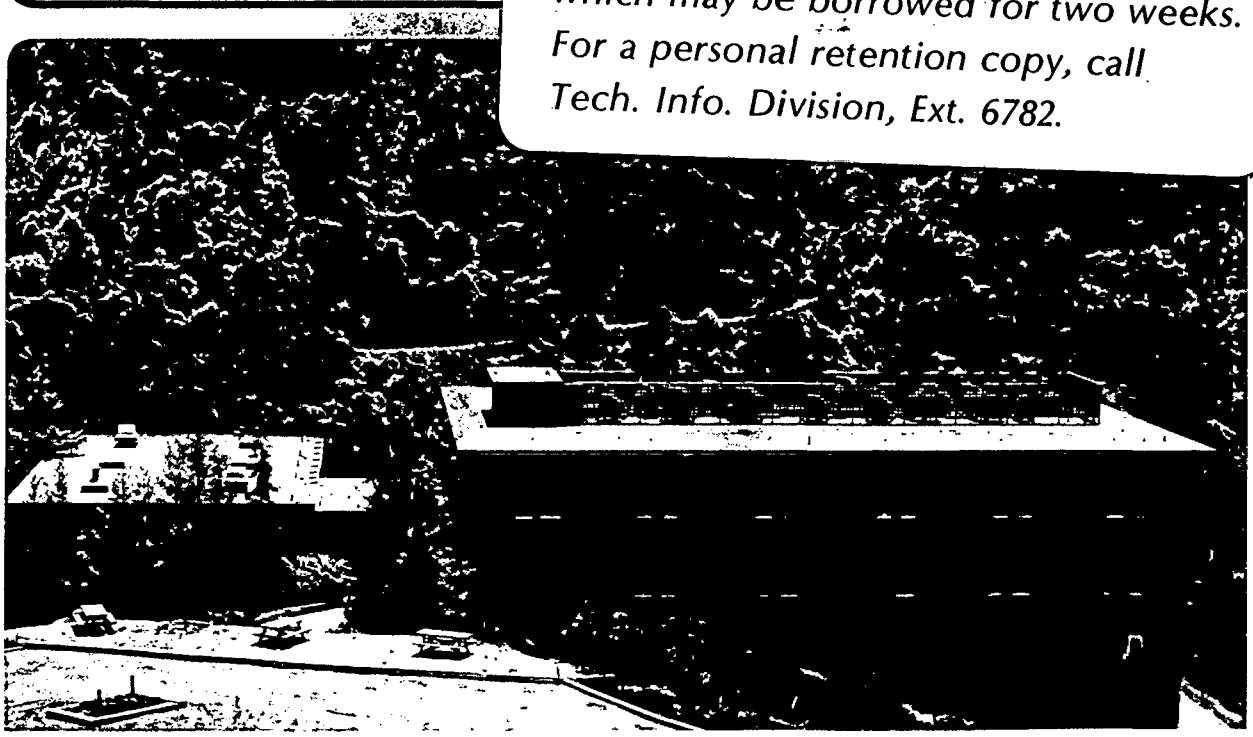

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LATH MARTENSITES IN LOW CARBON STEELS

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ABSTRACT

The morphology and crystallography of lath martensite in low and medium carbon steels have been studied by transmission electron microscopy and diffraction. The steels have microduplex structures of dislocated lath martensite ($a < b \ll c$) with fairly straight boundaries and continuous interlath thin films of retained austenite. Stacks of laths (i.e., single crystals of martensite) form the packets which are derived from different {111} transformation variants of austenite. Microdiffraction experiments directly allow the determination of the orientation relationships between austenite and martensite. Relative orientations of adjacent individual laths cluster about common orientations from small to large angular differences all around a common <110>M direction. The overall microstructure and orientations result from minimization of the total strain and shape deformation. Considerable accommodation occurs by deformation of laths (sometimes twinned) and austenite (sometimes tripped to twin martensite). In the meantime, microchemical analyses have shown considerable carbon segregation to the martensite-austenite interface.

INTRODUCTION

It is well-known that martensitic ferrous alloys have two types of morphologies, viz., twinned plates (extensively studied, especially from a crystallographic viewpoint) and dislocated laths (less extensively studied crystallographically), e.g. references 1-9. Recent studies of a range of experimentally developed medium and conventional low carbon steels have suggested that laths in the packets adopt orienations as a result of successive shears such as to minimize the shape deformation (analogous to the successive faulting shears in fcc $\rightleftarrows$ hcp martensite)(10,11). In the meantime,

microchemical analyses have shown considerable carbon segregation occurs to the martensite-austenite interfaces (which limits their mobility) (12,13) and that the austenite is stabilized by a combination of factors (carbon enrichment, immobile interfaces and mechanical stabilization) even though Ms can be well above 200°C. This paper is a continuation of the earlier studies (10-12). Attention has been focussed on obtaining more precise crystallographic data utilizing microdiffraction on the retained austenite and the stacks of laths in the packets.

Experimental Techniques. - Many experimental alloys have been used in this study. Those to be reported on here are low alloy-low carbon steels with high Ms temperatures (Table I).

Alloy No.	Composition (wt.%)	Ms°C
1	0.10C + 2Si	450
2	0.17C + 2Si	420
3	0.3C + 3Cr + 2Mn + 0.5Mo	340
4	0.3C + 3Cr + 2Ni + 0.5Mo	330

The homogenized specimens were high temperature austenitized (1100°C) and then quenched to produce lath martensite. Further details of the heat treatments and the preparation techniques can be found elsewhere (16).

RESULTS AND DISCUSSION

A. Morphology and Substructure. The microstructures consist of irregularly shaped packets of laths in which each packet is about 20-35μm in size and is composed of many laths of individual martensite crystals (with size ~ 0.5 x 5.0 x 20μm). Several packets can exist within the prior austenite grains. Laths have fairly straight

boundaries and are aligned near to the {111} habit plane variant of austenite (10,11). As is now firmly established, these heavily dislocated laths are separated by continuous thin films (~ 200Å) of untransformed, but highly deformed (12) austenite (retained austenite).

B. Crystallography. In the present alloy systems, it is commonly observed that parallel laths do not have the same orientation and that laths belonging to different poles are arranged in parallel with the common <110>M axis perpendicular to their broad faces in a single packet. Such a situation is shown in figure 1. Here in (b) several patterns corresponding to martensite ($[\bar{1}2\bar{1}]$, $[\bar{3}\bar{1}1]$, $[0\bar{1}1]$, $[\bar{1}11]$, $[\bar{5}\bar{1}1]$) and one corresponding to austenite ($[\bar{1}2\bar{1}]$) are superimposed. This result was also checked by dark field (DF) imaging (d,e and f). When the reflections belonging to different individual zones were used for the analysis, all laths corresponding to these variants reverse contrast. However, when the common 110 reflection is used, all the variants change contrast, including the retained austenite (whose 111 reflection is superimposed with this common (110)M). The important fact here is that there are many variants of differently oriented laths present within the same packet, but all having a common [110]M direction.

The microdiffraction data do not support the hypothesis given in ICOMAT '79 (11) that successive laths are "rotated" to produce orientation shifts adding up to 180°. Rather, there are only slight misorientations (1-10°) between parallel laths and they cluster about common <111> zone axes. In addition, for example figure 1, it is now shown that there are different stacks of parallel laths in a packet irregularly oriented with different zone axes. The question still remains whether laths, e.g., in a particular stack, are in the same <111>M orientation or in different

<111>M orientations. The uniqueness of this distinction requires very careful and tedious experiments involving microdiffraction, tilting and dark field imaging.

Controlled tilting experiments were carried out to identify the exact change in the orientation between adjacent stacks in a packet. One example, shown in figure 2 (in which the microdiffraction patterns were taken from the individual laths 1 through 19) reveal that there is a cyclic pattern of abrupt changes in the orientation from <111> to <001>, then back to <111>, and so on. This result has also been reported by others (5,7). By tilting the foil to about 35°, it was proved that those laths previously having <111>M orientation are now near the same <113> zone axis, and correspondingly those <001> laths are near the same <112> zone axis. These experiments show that, within the same packet of martensite, there are many stacks of laths corresponding to the same pole, e.g., $[\bar{1}11]$M, and there are only slight misorientations of a few degrees between the adjacent laths within a particular stack, and that the stacks themselves are separated by laths having a different orientation. The abrupt change between the poles, e.g., from $[\bar{1}11]$M to [001]M (or $[\bar{1}13]$M to $[\bar{1}12]$M) can be explained on the basis of the orientation relationship between the retained austenite and adjacent martensite laths. The small change in alignment around the [110]M direction will help to minimize the strain created during the γ to α' transformation.

C. <u>Orientation Relationships Between Austenite and Martensite</u>. The interpretations of the triple <111>M1/<110>A/<001>M2 diffraction patterns such as the one shown in figure 3h, were based on (14,15,17) conventional selected area electron diffraction. There are, however, only small angles involved in distinguishing between these orientation relationships(O-R), i.e., 5.26° between Kurdjumov-Sachs

(K-S) and Nishiyama-Wassermann (N-W), and a minimum of only 2.5° between Greninger-Troiano (G-T) and any of the K-S or N-W variants. Therefore, in order to unambiguously determine the correct O-R, one has to obtain isolated patterns from retained austenite and martensite crystals in either or both sides of the same lath boundary.

The experiment shown in figure 3 involves a region giving this frequently observed triple diffraction pattern. By using a small electron probe, microdiffraction patterns were taken (before and after 30° tilting) from the corresponding regions in austenite and martensite. From the analyses made in figure 3, it can be seen that the OR's are close to K-S and N-W. Within the reproducibility of the tilting experiment, however, the OR's are actually much closer to the G-T. The result of many experiments conducted by microdiffraction this way show that the OR's actually scatter between K-S, G-T, and N-W, but with G-T being the most frequently observed. The reason why so many orientation relationships may exist within a given packet may be attributed to the increased variants available for the laths during nucleation with {011}M//{111}A common boundaries in a packet (10,11).

D. Chemical Analysis. Considerable effort has been placed on determining the composition of the austenite and martensite phases. X-ray microanalyses show no substitutional element partitioning. However, lattice imaging, convergent beam electron diffraction and direct spectroscopy using field ion microscopy-atom probe methods (12,13) show conclusive evidence for carbon partitioning. In addition the carbon content at the austenite-martensite interface can attain up to 10 at% (see figure 4). (These results have already been discussed with respect to austenite

stability and whether the transformation can be considered bainitic (12,13,18). Therefore, these points will not be repeated here.)

SUMMARY AND CONCLUSIONS

1. The microstructure of experimental low and medium carbon-low alloy steels consists of dislocated, autotempered lath martensite with thin film retained austenite at the boundaries. The stacks of laths with different {111}A transformation variants form packets.

2. Adjacent laths in a particular packet have one <110>M direction in common and may be rotated about this axis with small angular ($1 < \phi < 10^{\circ}$) or large angular ($\phi > 20^{\circ}$), misorientations.

3. The orientation relationship between austenite and martensite all lie between K-S and N-W, but cluster in the intermediate G-T with highest frequency.

4. It may be hypothesized that upon quenching, several laths nucleate as discrete units in different parts of the preaustenite grain. Those which nucleate first establish the austenite variant, i.e., a particular {111} A, followed by other laths as they nucleate and grow one after another to form the packets. Retained austenite is then trapped at the boundaries. The relative orientation between adjacent laths, the orientation relation between austenite and martensite, the packet orientations, the structural dislocations, both in martensite and austenite, and the rare microstructural twins within laths are all arranged as a result of transformation and post transformation deformation to minimize the total strains. In rare cases, the austenite itself may "trip" to twinned martensite (16).

5. Previous work showed considerable carbon partitioning between austenite and martensite with higher carbon levels at the interface. This result indicates the importance of carbon on the interface mobility during the growth of the laths (a point previously suggested by Shoen et al. (19) who measured the growth kinetics).

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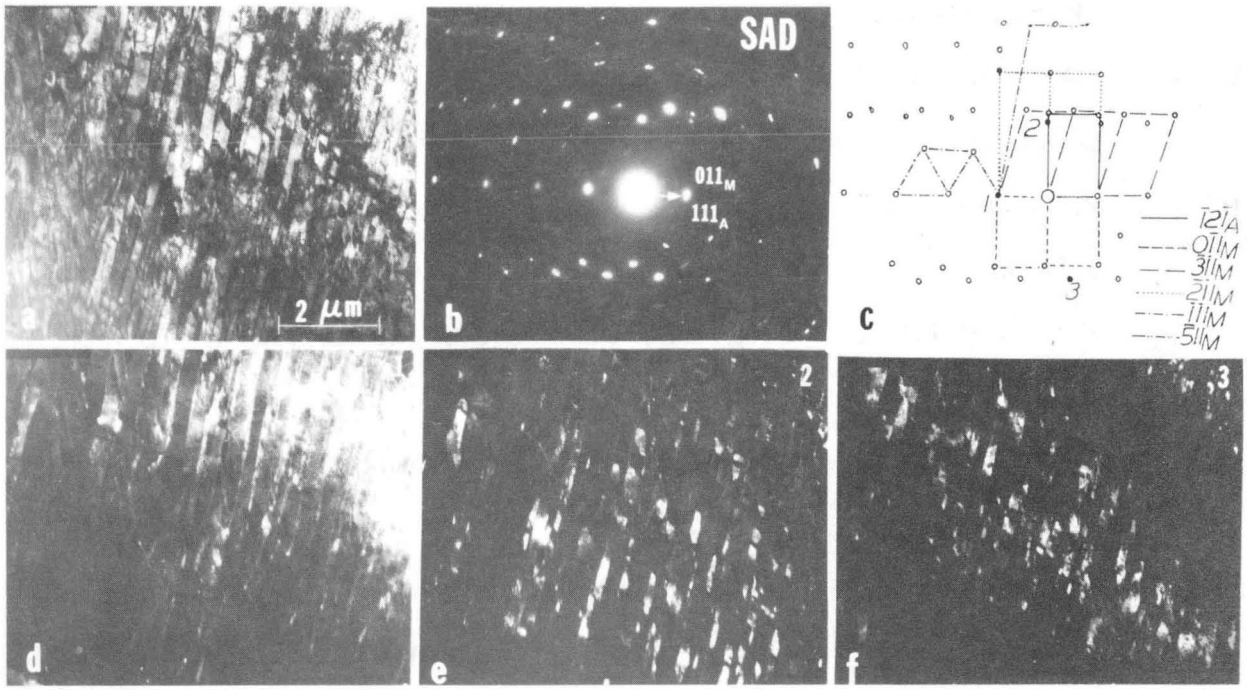

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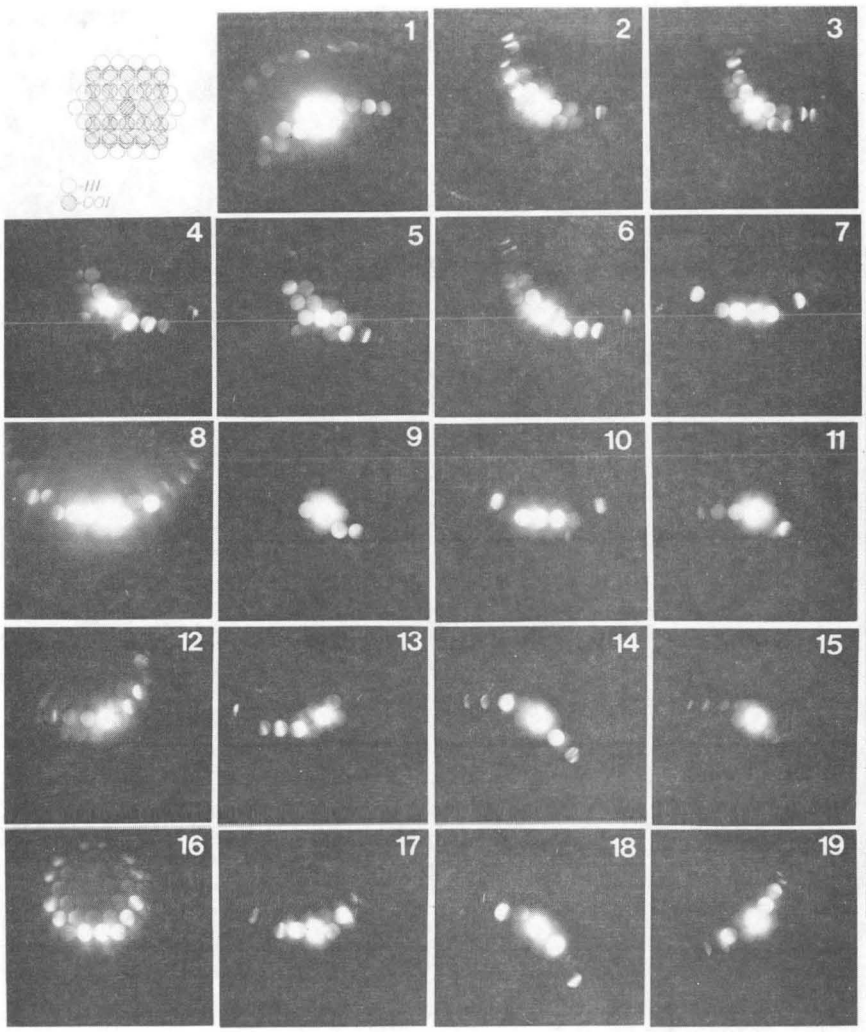
Fig. 1. Bright field (a) and selected area diffraction pattern taken covering many laths in (a). The orientations are sketched in c; dark field images shown in c,d,e from reflections 1,2,3 respectively.



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Figs. 2a,b. The microdiffraction patterns (1-19) in (a) were taken from the corresponding areas in BF image (b) before tilting.

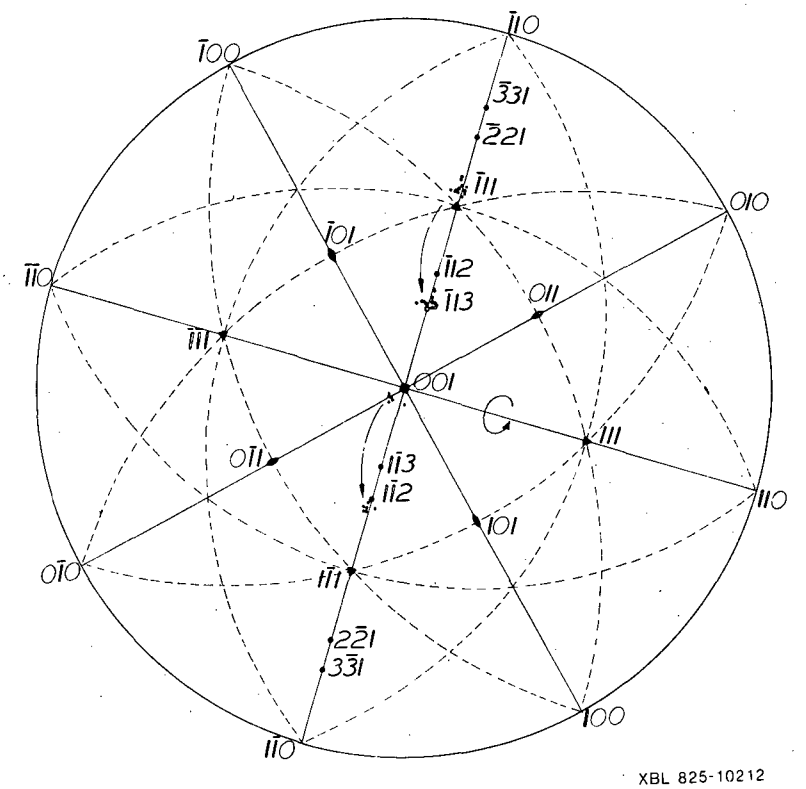
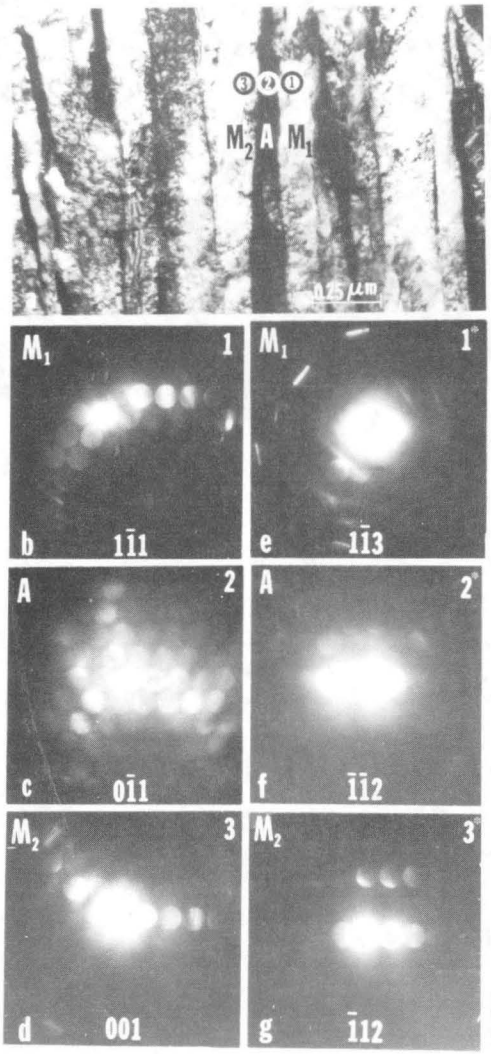


Fig. 2c. Stereographic projection analysis of the tilting experiment.



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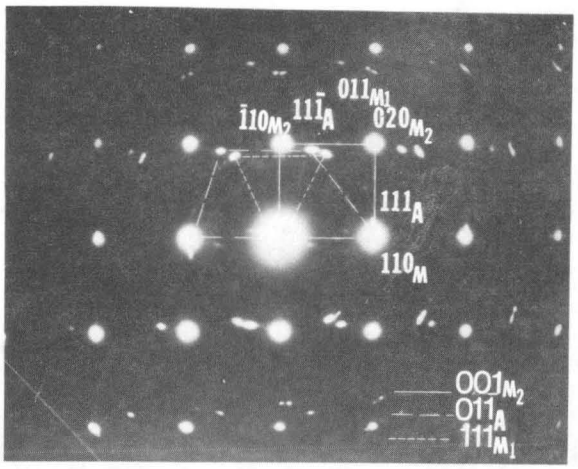

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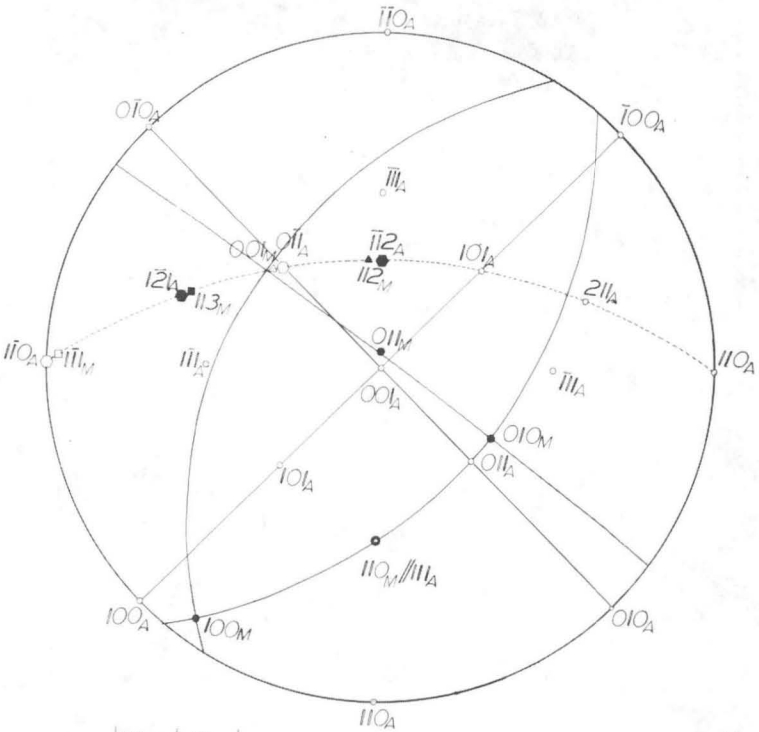

	Before Tilt	After Tilt
Martesite-1	□	■
Austenite	○	●
Martensite2	△	▲

$\left\{ \begin{array}{l} (110)_M \,//\, (110)_A \\ [1\bar{1}1]_M \text{ 2.5° off } [0\bar{1}1]_A \\ [001]_M \text{ 2.85° off } [0\bar{1}1]_A \end{array} \right\}$ OR $\left\{ \begin{array}{l} (110)_M \,//\, (110)_A \\ [1\bar{1}3]_M \text{ 2.00° off } [\bar{1}\bar{1}2]_A \\ [\bar{1}12]_M \text{ 3.50° off } [\bar{1}\bar{1}2]_A \end{array} \right\}$

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Fig. 3. Orientation relationship analysis. Patterns b-d taken before, and e-g taken after tilting from the corresponding areas in 3(a). The most commonly observed SAD pattern is shown in (h). The stereographic projection analysis is given in (i).

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