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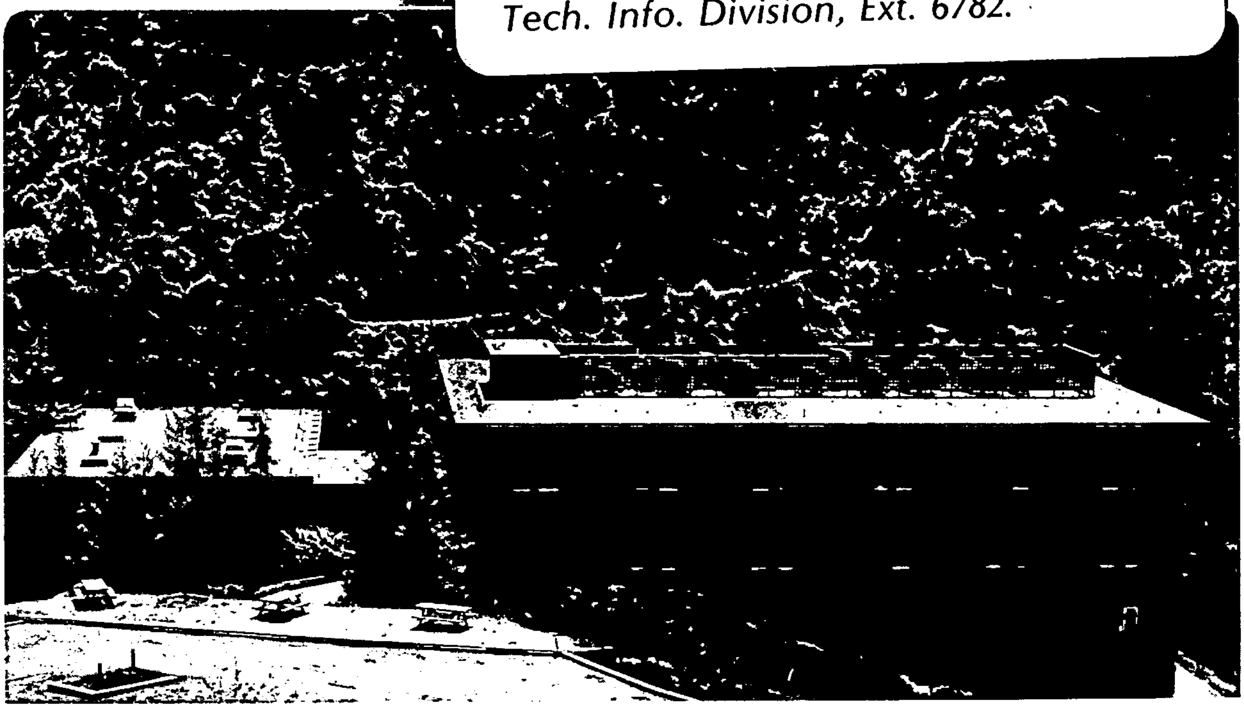

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LATH MARTENSITES IN CARBON STEELS--
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Summary

A detailed metallographic, crystallographic, and spectroscopic analysis of "lath martensites" formed by quenching low and medium carbon steels confirm an earlier suggestion that these structures are not strictly martensitic since carbon redistribution occurs during transformation. Perhaps a better description would be untransformed upper bainite with the interlath phase being carbon enriched austenite rather than carbide, as occurs in classic upper bainite. A summary of new crystallographic results using convergent beam electron diffraction is presented. In a given packet, the laths cluster around $<111>_{\alpha'}$ and $<100>_{\alpha'}$, but as reported earlier, other orientations also occur between them. The common axis between a particular bundle of laths is $<110>_{\alpha'}$. Rotations between bundles of laths and deformation of retained austenite may be important in minimizing the shape deformation.

Introduction

Over the past decade, considerable data on microstructure, crystallography, and properties have been accumulated in our alloy design program on low and medium carbon steels (see Refs. 1, 2), in which alloys are quenched from the austenite temperature range so as to produce dislocated packets of "lath martensite", in which the laths are almost always surrounded by films of austenite which have not transformed. These films are stable even to low temperature holding, but may transform at high stress concentrations (e.g., at the top of a growing crack in a K_{IC} test) and decompose to carbides on tempering (temper martensite embrittlement[3]). The question of the austenite stability is addressed in another paper at this conference[4] in which we also report new data on the carbon concentration at the martensite/austenite interface as well as in the martensite and austenite phases. Detailed experimental data from various techniques show that carbon partitioning must occur during the transformation of austenite.[4,5] Since this must occur by diffusion, then in a strict sense, the alloys are not ideally martensitic, which confirms our previous suggestion that a more appropriate characterization would be "untransformed upper bainite."[2] In the present work, a summary is given of recent detailed morphological and crystallographic studies of these structures.

Morphology and Crystallography

Figures 1 to 3 compare the morphology of directly quenched (lath martensite), tempered, and isothermally transformed ~ 0.3% C steels, respectively. In dark field, the morphologies are almost identical except that in Figure 1, the laths are surrounded by austenite. That is, as-quenched martensite can be converged into a microstructure almost exactly like upper bainite by tempering to allow the austenite to decompose by carbide nucleation at the α'/γ interfaces (Fig. 2). The fact that little or no intralath carbides occur in upper bainite indicates that the carbon content of bainitic ferrite is even less than in lath martensite (Fig. 4). This is to be expected since more time is available for complete carbon partitioning in the isothermal transformation situation. An actual carbon analysis across the microstructure similar to that of Figure 1 is shown in Figure 4. This distribution does not change appreciably from oil to brine quenched samples,[4] (note the cooling curve for an oil-quenched sample in Fig. 4a).

A detailed crystallographic study of lath martensites was presented at ICOMAT-79[2] at which time the need to redefine "lath martensites" was emphasized. It was also suggested that the relative orientation differences observed between laths in a packet was a consequence of minimization of the shape deformation. This was based on a model limited by the accuracy of conventional selected area electron diffraction.[2] Since Kikuchi patterns are too diffuse for analysis, we have subsequently carried out very detailed studies using convergent beam electron diffraction (CBED).[6] The new data is not exactly in agreement with the details of the rotation model proposed in 1979.[2] Figure 5 shows part of a detailed series of analyses of the orientation determination of laths in a single packet (i.e., one of the $\{111_{\gamma}\}$ variants) using CBED which uniquely defines the orientation, i.e., without the 180° ambiguity present in spot patterns, and Figure 6 shows a stereographic projection summarizing the data including controlled tilting about the $\langle 110\rangle_{\alpha'}$ axis common to all laths.[7] What these results show is that groups of 3 or 4 laths cluster around $\langle 111\rangle_{\alpha'}$. These are the most frequent observations, but some laths also cluster around $\langle 112\rangle_{\alpha'}$ and $\langle 113\rangle_{\alpha'}$. Thus, it may be that the shape deformation is not minimized by individual lath rotations, but by groups of several laths. Of course, the interlath austenite suffers considerable plastic deformation which itself accommodates transformation and deformation strains, as adjacent laths grow together, and deforms the retained austenite. The new observations reconfirm that retained austenite is not found between twin related laths[2]--which is another consequence of the shape accommodation.

The orientation relationships between austenite and martensite range from K-S through G-T to N-W, with G-T being the most frequent (Fig. 1). As suggested earlier, the reason why so many possibilities exist may be related to the choice of the most favorable nucleating condition and shape accommodation, i.e., the more potential variants, the better.

Wakasa and Wayman[8] also studied lath crystallography in carbon-free steels, but with low M_2 temperature. Since it is known that carbon is needed to stabilize retained austenite in steels with high M_s, it is not surprising that they did not find retained austenite and that lath orientations in a packet clustered around a single pole.

On the other hand, Eterashvili et al.[9], working on low C-low alloy steels, published almost identical results to those reported here.

Thus, it should be emphasized that the presence or absence of retained austenite is of fundamental significance in descriptions of the transformation. The detectability of such austenite requires very careful characterization using the most advanced experimental techniques now available.

Description of the Transformation; Conclusions

On cooling, the austenite of the nominal composition of the alloy transforms at M_s to form a lath, or initially a plate, with subsequent rejection of carbon across the γ/α' interface. As long as the M_s-M_f temperature range is high enough to allow carbon diffusion to occur (and for low and medium carbon steels, this is usually above 200°C), the martensite grows laterally and loses about 10% or more of its carbon content, Figure 4. Eventually, the carbon content at the interface can reach very high values, ~ 10% at.% which must seriously reduce the mobility of the α'/γ interface. At this stage, transformation stops leaving retained austenite with a carbon level ~ 3 at.%. At this carbon level, the retained austenite M_s is calculated to be ~ 150°C, so it must be stabilized by other mechanisms (see Fig. 4 of ref. 4). During this continuous cooling, the austenite does not have time to transform to carbide, but on subsequent tempering or isothermal annealing, the austenite can then decompose by carbide nucleation at the α'/γ interface where the carbon content is a maximum. Thus, the as-quenched state is that of laths surrounded by stabilized austenite, rather than carbide as in classic upper bainite (Fig. 3). It should be noted, however, that retained austenite has been identified in bainitic steels[10,11] and an example[10] is given in Figure 7.

A comparison of Figures 1, 2, 3, and 7 shows that a range of structures can be obtained from the transformation of austenite in low and medium carbon steels. These can be summarized as follows:

--α' laths with interlath carbide (upper bainite),

--α' laths with intralath carbides, retained austenite (lower bainite),

--α' laths with retained austenite (also autotempered intralath carbides),

--α' laths without austenite (unusual in carbon steels with $M_s > 250$°C).

Thus, depending on composition, carbon diffusion and partitioning, and the mobility of the γ/α' interfaces, which in turn depend on alloy content and heat treating conditions, many combinations of microstructure containing mixtures of the laths, austenite and inter/intralath carbide phases are possible.

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Figure Legends

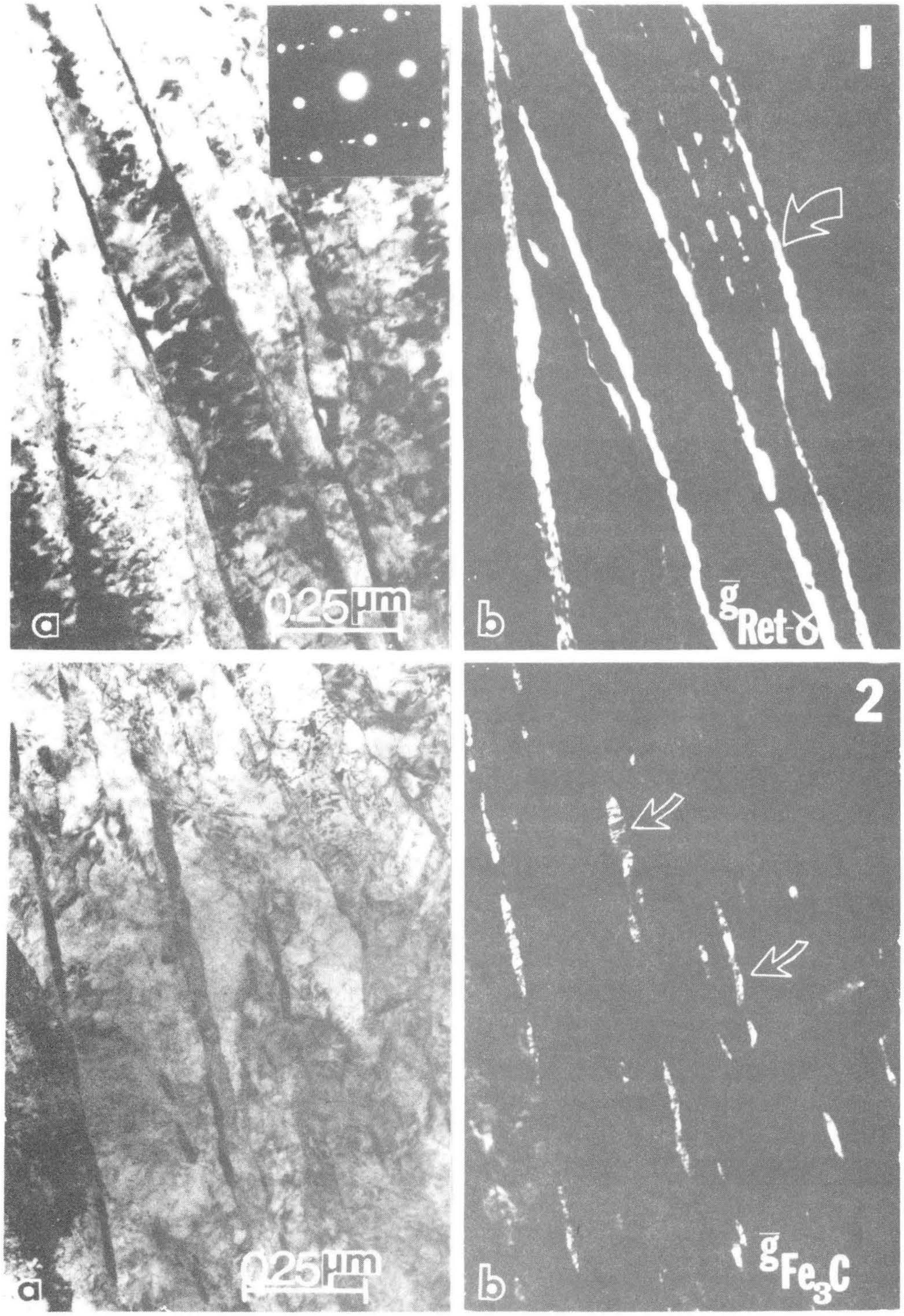
Figs. 1, 2, and 3. Comparison of the microstructures in 1) as-quenched lath martensitic, 2) tempered martensitic (T = 300°C), and 3) upper bainitic (T = 360°C) states. Fig. 1 - Fe/3Cr/2Mn/0.5Mo/0.3C, M_s = 320°C (note the characteristic $\langle 110\rangle_\gamma - \langle 111\rangle_{\alpha'} - \langle 100\rangle_{\alpha'}$ triplet SAD pattern corresponding to the K-S and N-W relations). Fig. 2. Fe/3Cr/2Ni/0.5Mo/0.3C, M_2 = 340°C. Fig. 3. Fe/4Cr/0.34C, M_s = 320°C.

Fig. 4. a) The cooling curve for oil quenched Fe/3Cr/2Mn/0.5Mo/0.3C samples (1mm dia.). b) Schematic diagram similar to an actual atom probe analysis[4]) shows the approximate carbon profile (with spikes corresponding to c clusters at defects) across the α'-γ duplex microstructure in the as-quenched sample (see Fig. 1).

Fig. 5. a) Bright field electron micrograph from the 0.3C steel showing the regions corresponding to the CBED patterns in b) taken from the individual α' laths before and after tilting.

Fig. 6. Stereographic projection summarizing the data of tilting experiments, such as Fig. 5.

Fig. 7. a) Bright field and b) dark field micrographs from Fe/2Si/0.4C steel (M_s = 370°C) to produce lower bainitic structure. DF micrograph ($\overline{g_{\bar{1}1\bar{1}_\gamma}}$) reveals S-shaped retained austenite along ferritic lath boundaries, in addition to intralath carbides (from Ref. 10).



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Figs. 1, 2

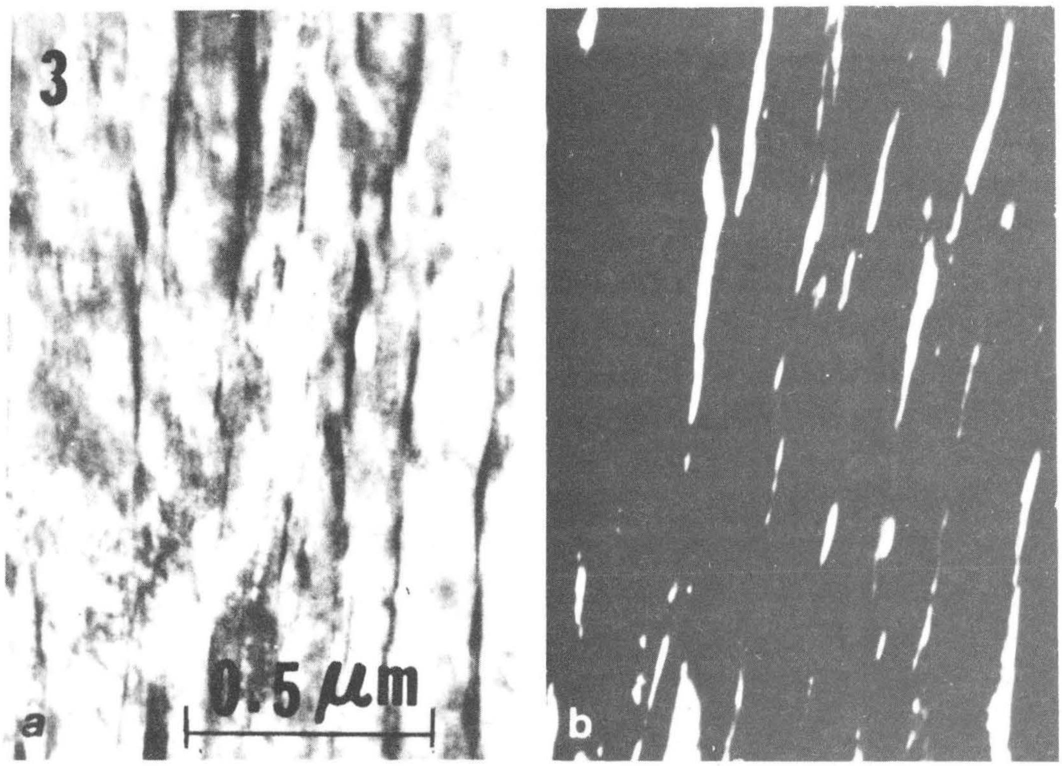

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Fig. 3

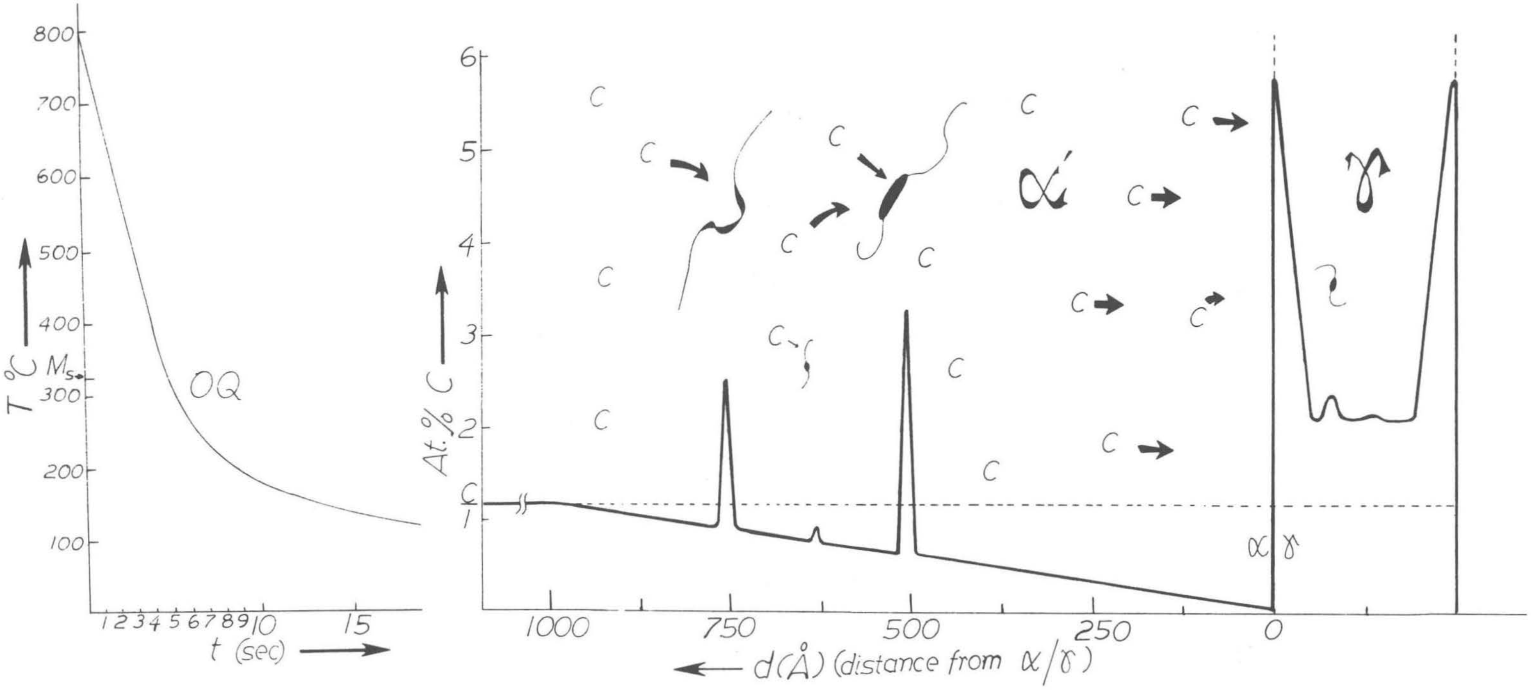

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Fig. 4

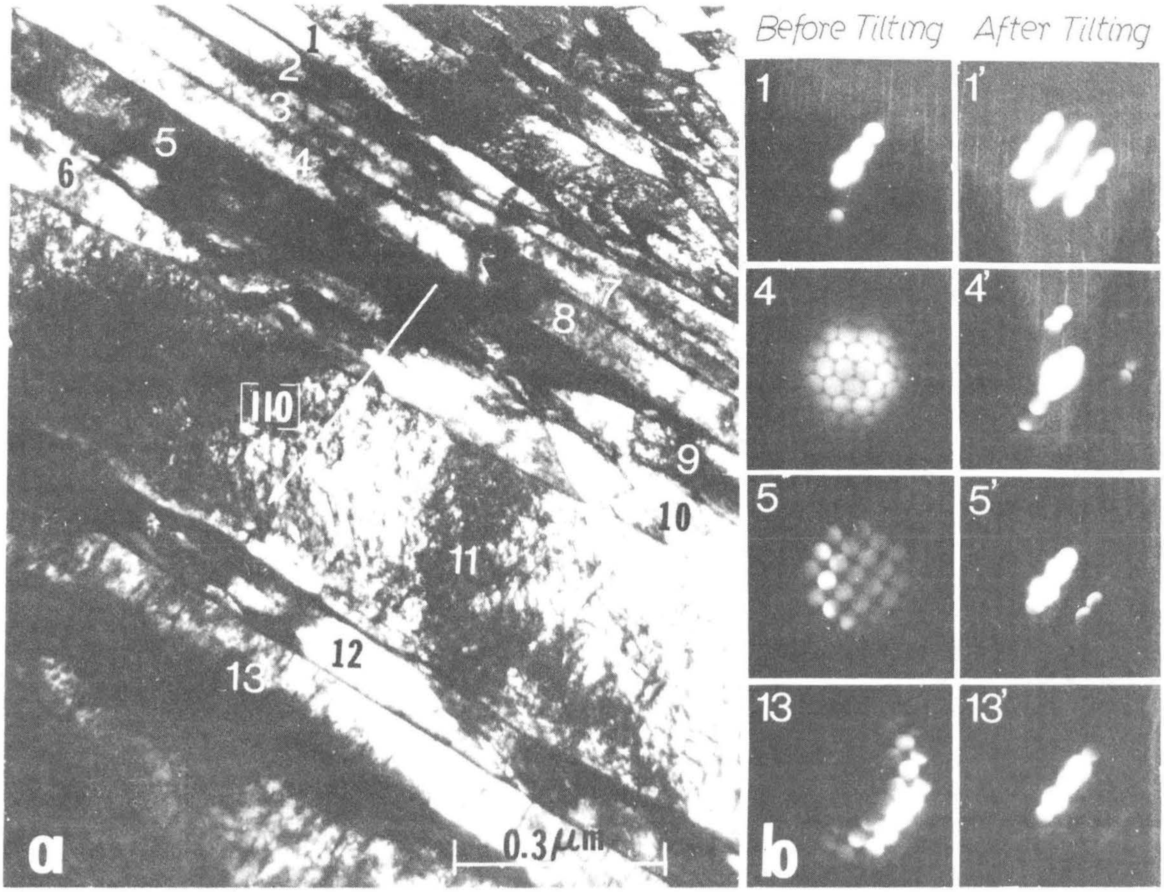

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Fig. 5

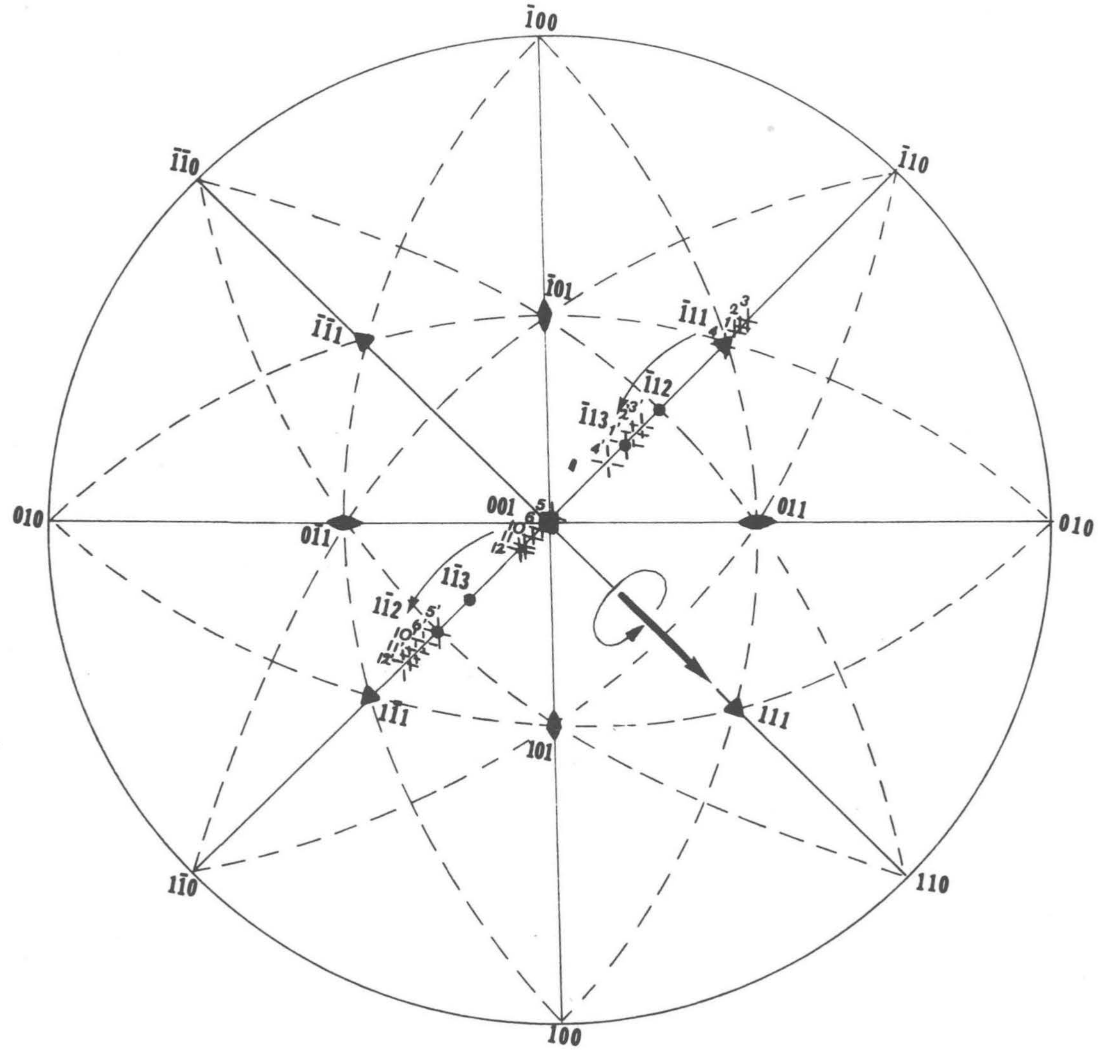

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Fig. 6

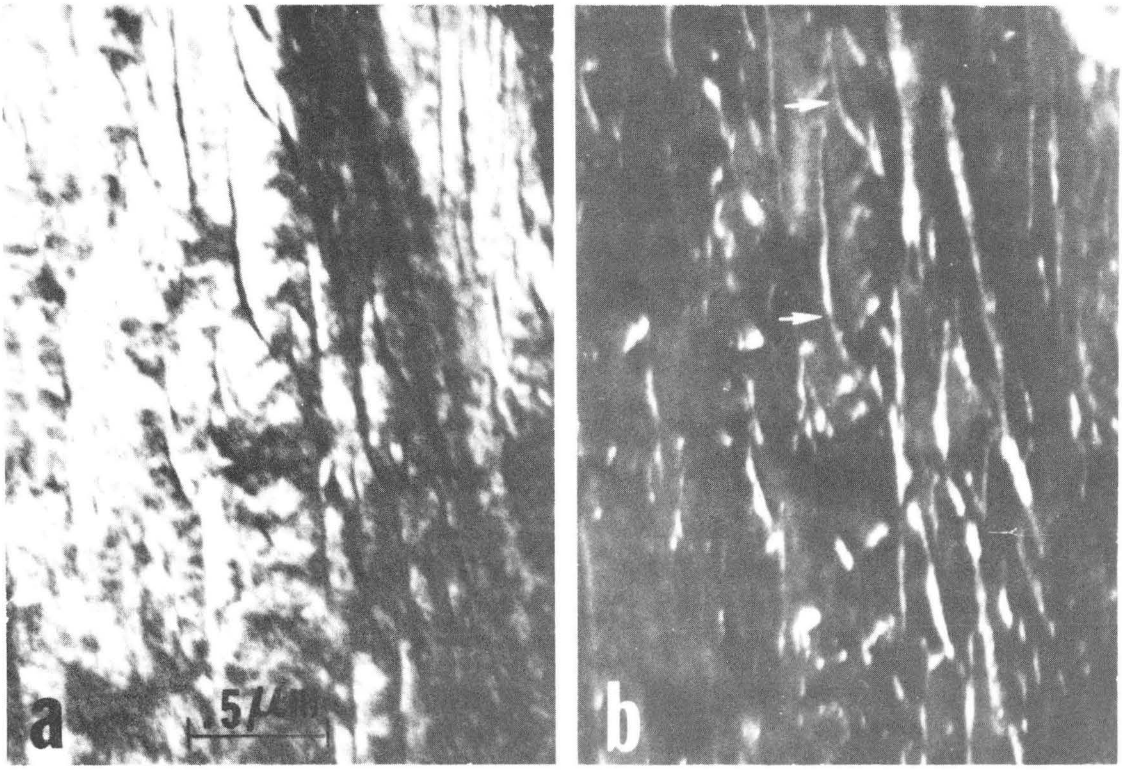

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Fig. 7

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