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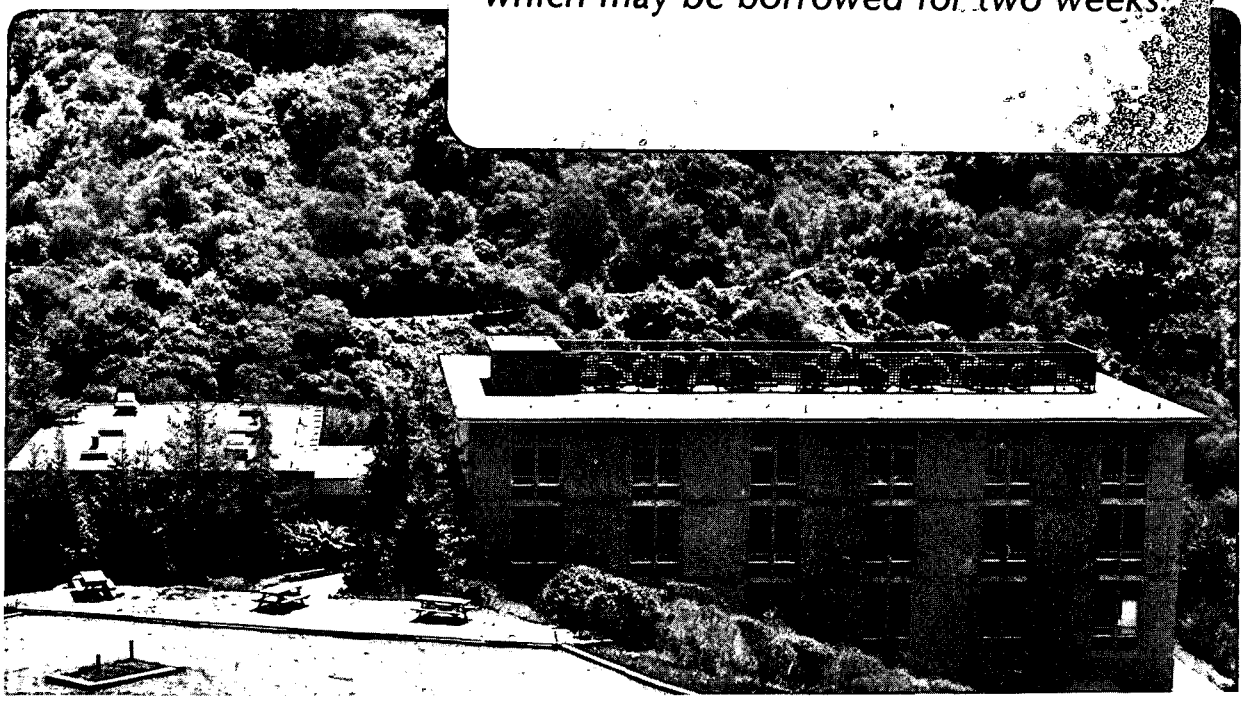

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LATH MARTENSITE AND BAINITE IN LOW ALLOY STEELS

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LATH MARTENSITE AND BAINITE IN LOW ALLOY STEELS

The morphology and crystallography of lath martensite and bainite have been studied by transmission electron microscopy, TEM, imaging and diffraction techniques. When quenched from the austenite phase region, the steels have microduplex structures of dislocated lath martensite ($a<b<<c$) with continuous interlath films of retained austenite. Bainites form when interrupted quenching is applied at temperatures around M_s to allow isothermal transformation to take place. The microstructure of lower bainite resembles that of lath martensite, where bainitic dislocated ferrite laths, which now contain unidirectional carbides, are surrounded by thin film retained austenite. In upper bainite, the boundaries are decorated by interlath carbides, a microstructure observed in martensite tempered at 240-400°C. In all cases laths form packets in which all the laths transform from a (111) variant of austenite.

Microdiffraction experiments reveal that bainitic ferrite laths are related to adjacent laths as well as the parent austenite through the same orientation relationships as faund in martensite in the same alloys. This strongly confirms the shear aspect of bainitic transformations. Bainitic carbides, identified as cementite, are found to have the Isaichev orientation relationship with ferrite, which supports the idea that cementite has transformed directly from austenite. The microstructural and crystallographic characteristics of martensite and bainites in low carbon steels are very similar which confirm that these microstructures are the products of a continuous range of phase transformations from austenite where shear is the basis, and the morphological features develop depending on the time allowed for carbon diffusion, and carbide precipitation.

Introduction

For many years both the martensite transformation and the bainite transformation have been widely used to obtain a variety of strength-toughness combinations in ferritic steels. However, due to inherent complexities of these non-equilibrium phase transformations, a complete understanding has not been achieved despite tremendous efforts made from the metallographic, kinetic, crystallographic and phenomenological theory viewpoints. In particular, steels containing relatively low concentrations of carbon ($C<0.3$ wt%) and other alloying elements (total <5 wt%), although frequently used for various engineering applications, have rarely been subjected to fundamental studies of martensite and bainite transformations, until recently. The high transformation temperatures ($M_s>200$°C) and high reaction rates of these steels especially render direct kinetic studies difficult. However, the advancements in diffraction and microanalysis techniques in TEM in recent years have stimulated intensive research into the microstructure and crystallography of lath martensite in various low alloy steels (1-4).

In parallel with an alloy design program to achieve high strength and high toughness experimental steels involving relatively simple compositions and heat treatments, the crystallography has been studied in detail to understand transformation characteristics of the structures developed (1,4,5). The steels have compositions from 0.1 to 0.4 wt% C, 2-4 wt% Cr, 1-2wt% Mn, and with or without Mo, Ni, (1,4-9). When quenched from the austenite phase field, the steels produce lath martensite with retained austenite in the form of thin films at the lath boundaries. The morphology and crystallography

of lath martensite, in relation to retained autenite, in these alloys are now well studied and documented (1, 4-8). As a consequence of these developments to achieve desired properties, the possibilities exist for bainitic steels with similar compositions as lath martensitic steels for thick section applications (9). In this paper, a summary will be given of the detailed TEM investigations of the crystallography of bainitic structures in refence to the similarities of the lath martensitic stuctures produced in these steels.

Because of the many features which are intermediate between the pearlite and the martensite transformations, the bainite transformation has been a subject of debate in many regards, in particular to its transformation mechanism. Two distinct theories have been proposed (2, 10-11): one of them considers the transformation to be an extension of the martensitic transformation to higher temperatures, analogous to autotempered martensite, i e, shear controlled transformation, and the other theory considers that the mec anism which governs both the pearlite reaction and Widmanstatten ferrite formation also applies to the transformation at lower temperature, i.e., bainite transformation is solely diffussion controlled. In this paper a high degree of similarities in the morphology and crystallography of upper and lower bainites with lath martensite has been found in the same alloys, which reinforces the theory that bainite transformation is indeed shear controlled in such alloys.

Experimental Techniques

The experimental alloy used was 0.3% C-3.0% Cr-2.0% Mn (in wt%) with an M_s temperature of 320 °C. The homogenized specimens were austenitized at 1100 °C and then either quenched in oil or isothermally transformed at 360 °C to produce martensite or bainite, respectively. Thin foils for TEM observations were made by electropolishing in a CH_3COOH (400 ml)-CrO_3 (75 gr)-H_2O (20 ml) solution at room temperature. Conventional imaging and microsdiffraction experiments were conducted in a Philips EM 400 TEM.

Results and Discussion

The crystallography of lath martensite in the Fe/Cr/C system, and in particular in 3.0% Cr-2.0% Mn-0.5% Mo-0.3% C steel, has been studied in detail (1, 4-8). It has been proposed that martensite forms by shear in a $(111)_A$, parallel to $(110)_M$, where A and M refer to austenite and martensite respectively. Strains build up in the surrounding austenite by the formation and the growth of the first lath and this promotes subsequent lath formation, while mechanically stabilizing the austenite trapped between the laths At the same time, the concentration of carbon at the A/M interfaces reaches a high value (~1 wt%) (e.g., ref. 5) During the successive formation, adjacent laths arrange their relative orientations with respect to each other such that the overall transformation strains are minimized (7-8)

Parallel laths form a bundle with their $\langle 110 \rangle_M$ axis perpendicular to the lath boundaries in a martensite packet where all the laths are from the same $\{111\}_A$ variant (Figure 1) The relative orientations of adjacent laths which have resulted form the strain minimization are observed in various degrees of rotation, $\emptyset$, about the common $\langle 110 \rangle_M$ (4-8) O can be a small angle ($0^o < \emptyset < 10^o$), large angle ($\emptyset > 20^o$), or in twin relation. Thus the orientation relationship between the retained austenite and martensite varies from Kurdjumov-Sachs (K-S) (12) relationship through the Greninger-Troiano (G-T) (13) relationship to the Nishiyama-Wassermann (N-W) (14-15) relationship across the packet

As far as the austenite and ferrite phases are concerned, the crystallographic characteristics observed in the lower bainitic structure of the present steel are mostly in close agreement with the above results obtained the lath martensite. For example, $\langle 110 \rangle_B$ direction is perpendicular to the lath boundaries in lower bainite (Figure 2) even if the relative lath orientation is changed, and by the presence of retained austenite, this direction has been shown to be parallel to a $\langle 111 \rangle_A$ direction Therefore, a $(111)_A$ type habit plane is also present in bainite. (Here B refers to "bainitic ferrite".)

Three types of arrangements of adjacent laths have been revealed by microdiffrac-

tion experiments. These are: (i) clustering of orientations around a particular pole; (ii) abrupt change in the orientation from $\langle 111 \rangle_B$ to $\langle 110 \rangle_B$; and (iii) twin related laths. As shown in Figure 2, where, in addition to a number of laths in different zone axis orientations, retained austenite is also present. Therefore, cases (i) and (ii) are the same as was observed in lath martensitic structures. However, twin related laths in lower bainite are the $(112)_B$ type, while in the martensite, the $(110)_M$ type. This arrangement can still be attributed to strain minimization during martensite formation. This can be discerned from the fact that no carbide was observed at these boundaries, suggesting that there was no austenite present originally at these sites.

Although these results have clearly shown the similarity between the crystallography of bainitic ferrrite and retained austenite and that of lath martensite and retained austenite, this is not exactly the case when considering the crystallography and morphology of cementite in bainite and in tempered or autotempered martensite. The orientation relationship between cementite and bainitic ferrite is the Isaichev relationship (15), namely: $(111)_B // (010)_C$, and $(011)_B // (103)_C$, instead of the Bagaryatskii orientation relationship (16) between cementite and lath martensite: $(110)_B // (100)_C$, $(111)_B // (010)_C$, and $(112)_B // (001)_C$. However these relationships are not totally different form each other. The Bagaryatskii relationship can be obtained by a rotation around the $[010]_C$ axis of approximately 4^0. The principal axes of the cementite phase are oriented so closely in the direction that these two orientation relationships have been regarded as one relationship which is typical of tempered martensite in some previous studies (18). However, this small difference between the orientation relationships can not be ignored. When the Isaichev relationship is obeyed, the $(103)_C$ plane is parallel to the $(011)_B$ plane, and which in turn is parallel to lath boundaries. As discussed above, $(011)_B$ is very likely to be the habit plane of the bainite transformation and hence it is parallel to a $(111)_A$ variant. Then there is a parallelism between these phases, namely: $(111)_A // (011)_B // (103)_C$. This is indeed observed in Figure 2. If this is the case, then the K-S orientation relationship is obeyed between bainitic ferrite and retained austenite, namely: $(111)_A // (011)_B$, $(110)_A // (111)_B$, and $(112)_A // (211)_B$. Accounting for this, another parallelism between these three phases is derived, namely: $(110)_A // (111)_F // (010)_C$. These two parallelisms indicate that the closepacked planes and the closepacked directions, respectively, in austenite, ferrite, and camentite, are parallel to each other. Assuming these parallelisms, the correspondence between austenite, ferrite, and cementite is illustrated in Figure 2 d. From this stereogram, a new O-R can be now be established, namely: $(554)_A // (100)_C$, $(110)_A // (010)_C$, and $(225)_A // (001)_C$, which corresponds to the Pitch relationship (19). This was first found between austenite and pre-eutectoid cementite in hypereutectoid C-Mn steel (19). Therefore, it can be concluded here that bainitic cementite was initially formed from austenite. The same conclusion was drawn earlier for 2% Si steel and 12% Ni-4% Co steel (20).

The above analysis is consistent with the fact that there are differences in the habit planes and morphology of cementite between bainite and martensite, the former showing the Isaichev and the later Bagaryatskii orientation relationships. Previous studies (e.g., References 11 and 18) as well as the present work indicates that the habit plane of cementite in lower bainite is near $(112)_B$, and that the morphology of cementite is unidirectional (9, 18, 20-21) i.e. one habit plane in a lath. In tempered martensite, on the other hand, cementite has $\{110\}_M$ habit and exhibits a Widmanstatten morphology, i.e., two or three habit planes in a lath (2,7-8, 22-25).

These studies, as summarized in Table-I, suggests that the ferrite matrix froms by a shear transformation, and the crystallography of bainitic cementite indicates that the carbides precipitate directly from austenite involving diffusion. Therefore, the bainite transformation must combine both shear and diffusion controlled behavior, as pointed out, e.g., by Cohen and Wayman (26).

Summary and Conclusions

It has already been established in previous research (e.g., ref. 5) that lath martensitic transformation in these alloys is accompanied by carbon diffusion during cooling, probably at the A/M interface. The degree of carbon partitioning has been established, viz., while the average composition of retained austenite increases to 1.0 wt%, it is >2.0 wt% at the austenite/martensite interface. It is now known that retained austenite is also present in lower bainite and the morphological and crytallographical features in lower bainite resemble that of martensite. It can be concluded, then, that the mechanism of these phase transformations are similar. In all cases, austenite to ferrite (martensitic or bainitic) transformation involves shear, but the rate of the transformation is determined by carbon diffusion at the A/M interface, which, within the available time, either stabilizes austenite (in lath martensite, and to a certain extend in lower bainite) or forms autotempered carbides (lath martensite), unidirectional carbides (in lower bainite), or interlath carbides (in tempered martensite and in upper bainite) as is schematically shown in Figure 3. A summary of the comparison of the crystallographic and morphological features in these structures is given in Table-I.

Table-I: Comparison of Lath Structures in Low Carbon-Low Alloy Steels
Fe/3 Cr/ 2 Mn/ X C : $0.05 < C < 0.4$, $M_s > 200\ ^oC$

Feature	Martensite	Bainite	
		Lower	Upper
Laths in Packets	high dislocation density laths	dislocated laths	dislocated laths
	dislocation density decreases --------►		
Common Direction	lath faces $\langle 110 \rangle_M$	lath faces $\langle 110 \rangle_B$	irregular
Interlath Region	continuous retained austenite (quenched) or stringers of M_3C (TME)	non-continuous retained austenite (short isothermal) or carbides (long isothermal)	some retained austenite, but mostly carbides as in TME*
A/F Interface	C enriched	some C enrichement	probably none
Austenite Stability	can be very stable	not very stable	not very stable
Interface plane	$(110)_M // (111)_A$	$(110)_B // (111)_A$	---
A/F orientation relationships	K-S to G-T to N-W	K-S to G-T to N-W	---
Twin Relation of Laths	$(110)_M$ plane	$(112)_F$ plane	not observed
A/F microscopic habit	$(111)_A$	$(111)_A$	inferred
Carbides and Orientation Relationship	Widmanstatten Bagaryatskii $(110)_M // (100)_C$ $(111)_M // (010)_C$ $(112)_M // (001)_C$	unidirectional Isaichev $(111)_B // (010)_C$ $(011)_B // (103)_C$	at boundaries ---
Mechanism of Transformation	shear + interface C diffn and autotempered carbide pptn	shear + C diffusion interlath carbide precipitation	shear + C diffusion interlath carbides

* TME means tempered martensite embrittled condition (27)

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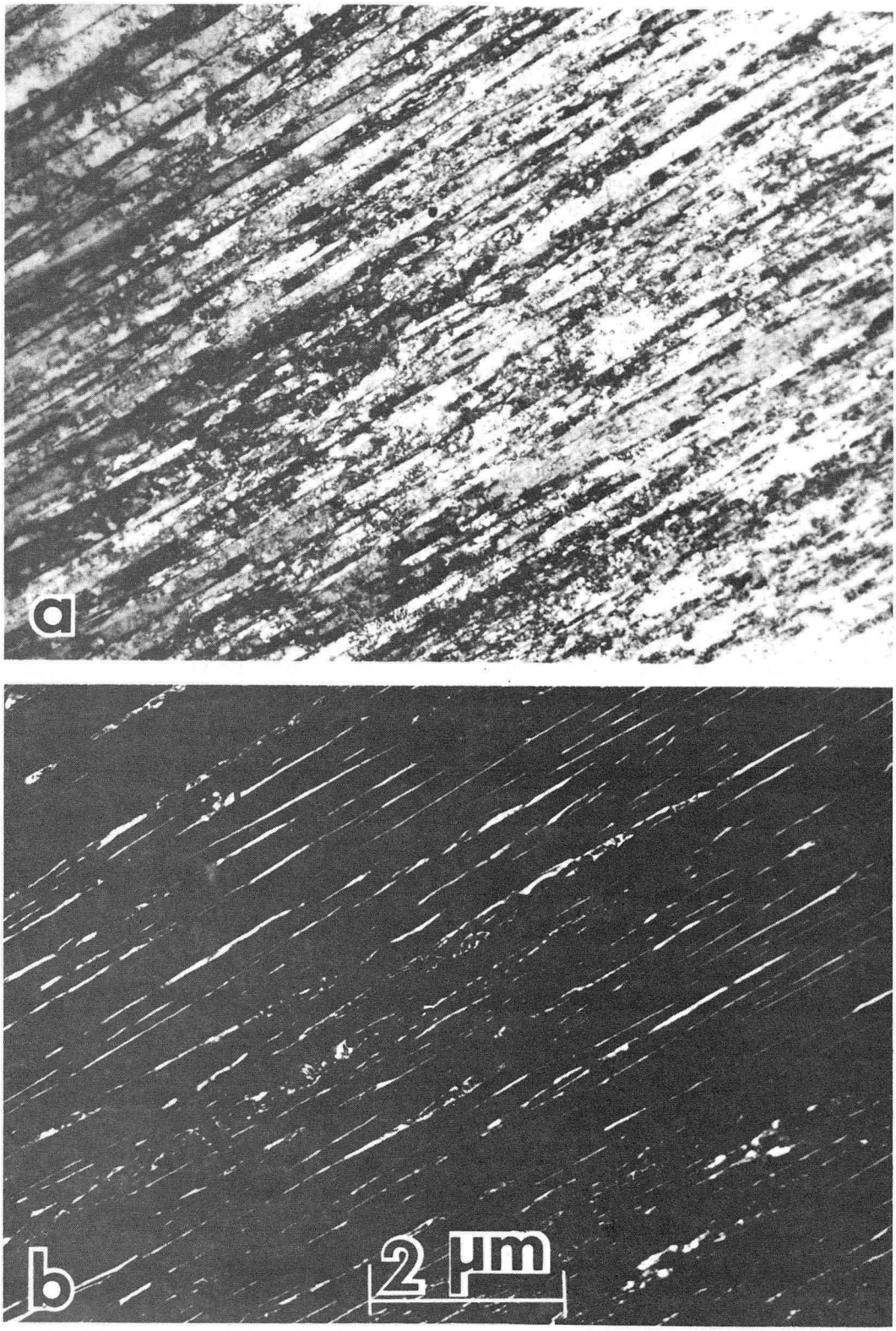

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Fig. 1

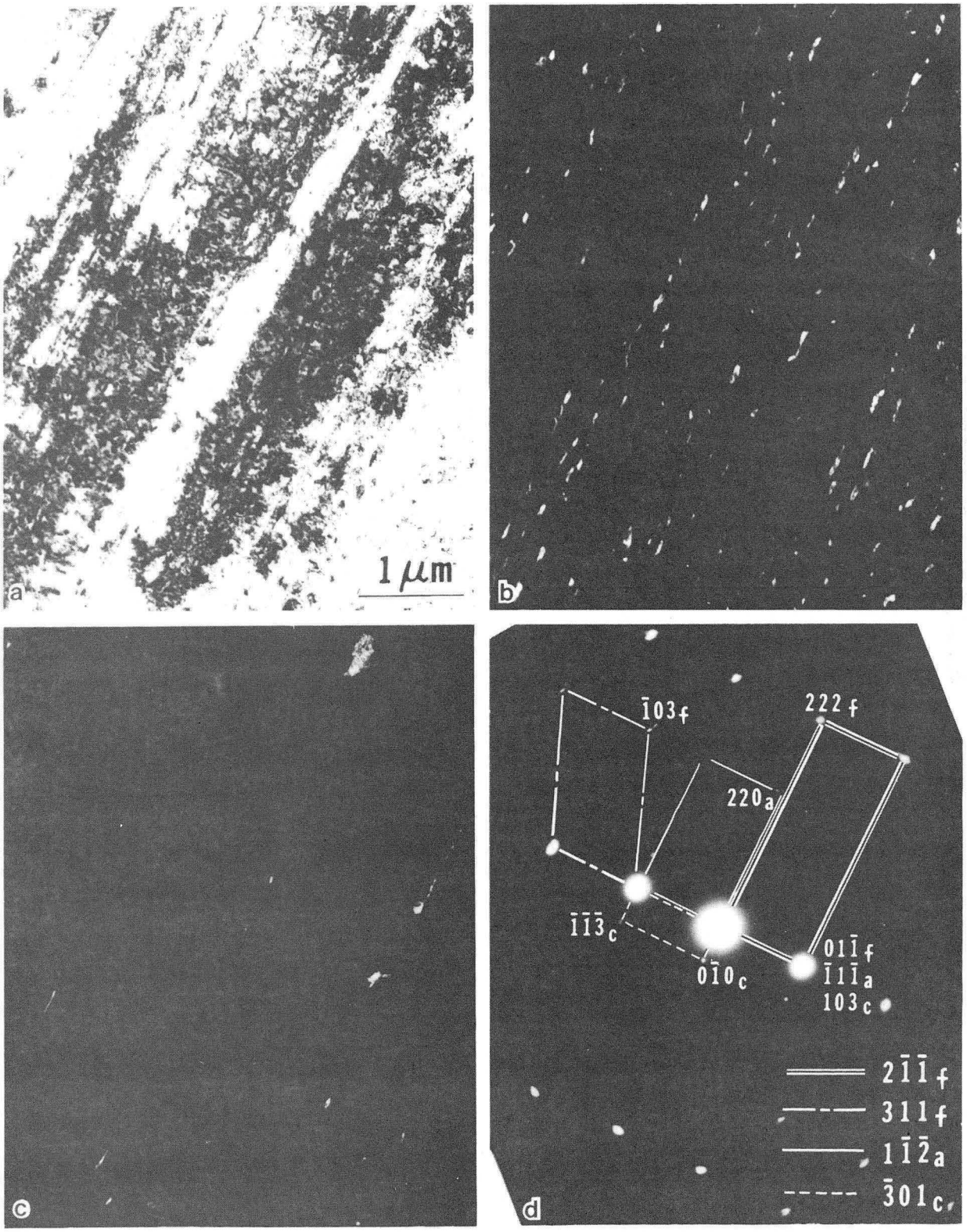

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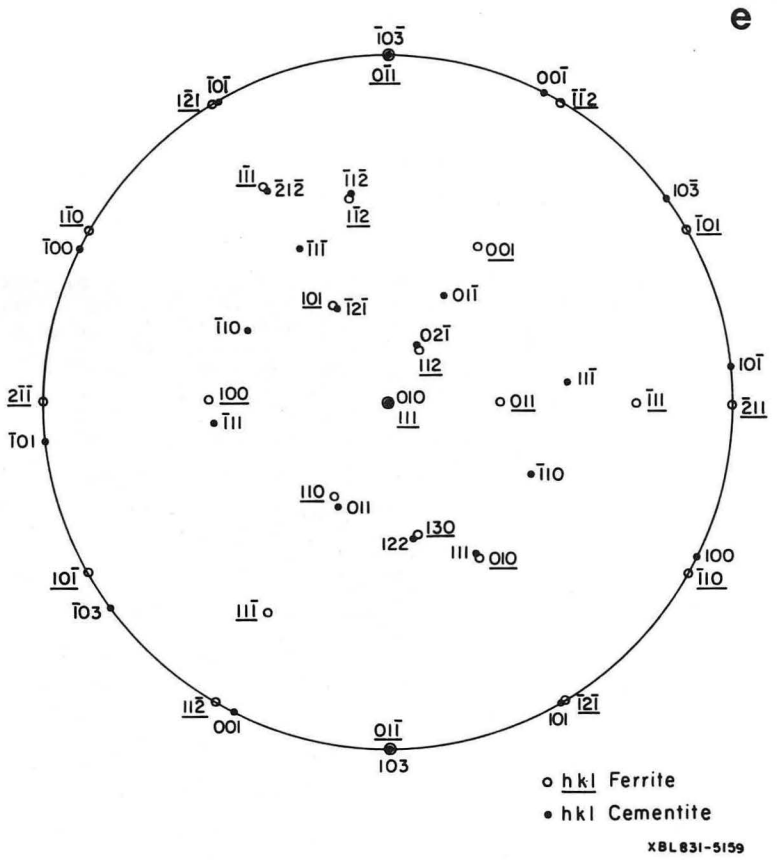
Fig. 2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hkl Cementite
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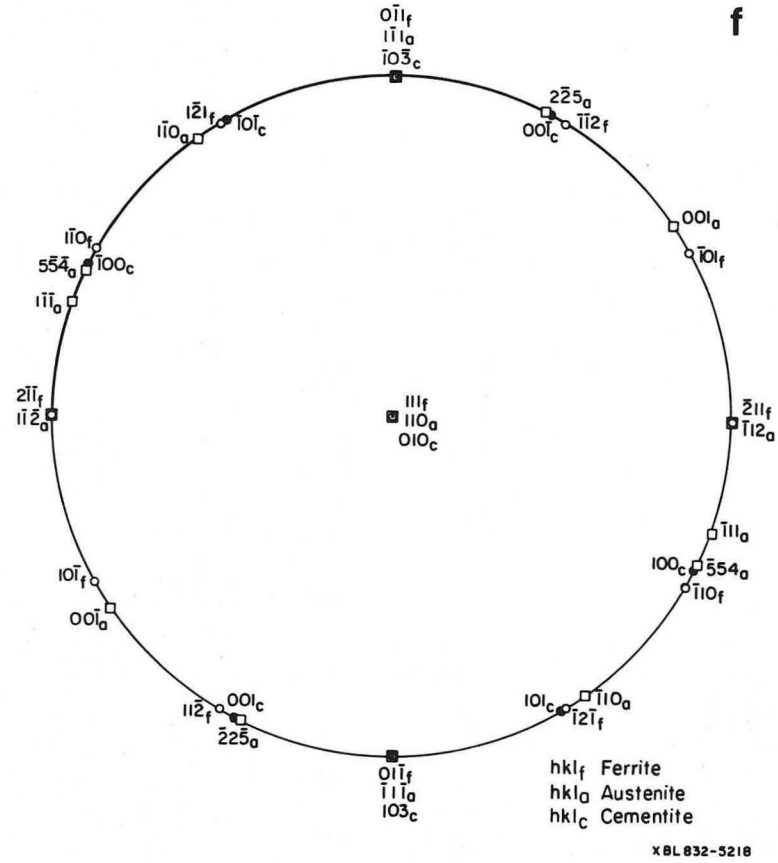
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hkl$_c$ Cementite
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Fig. 2

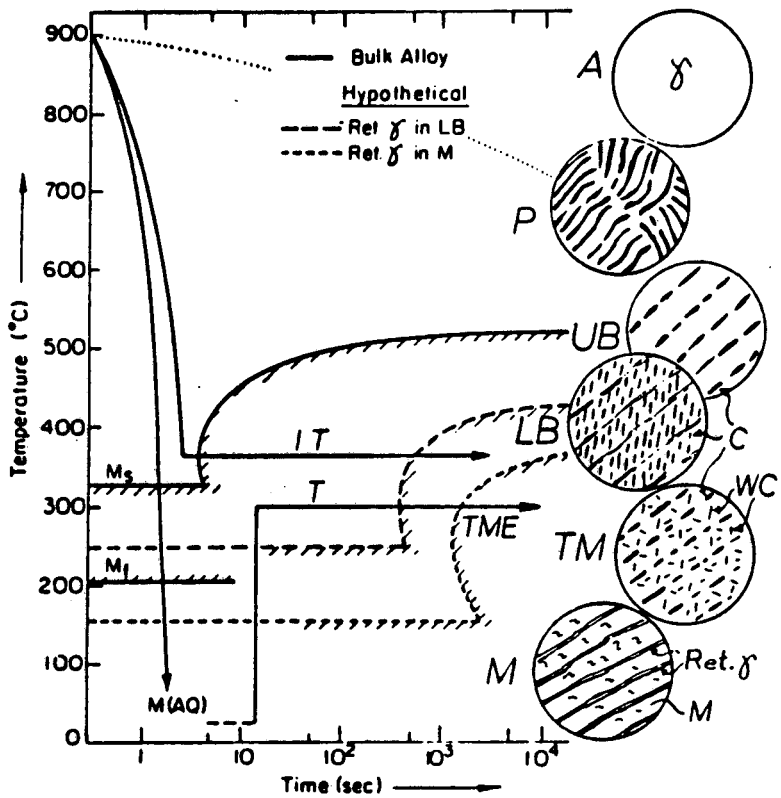

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Fig. 3

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