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OF DUAL PHASE Fe/Si/C STEELS

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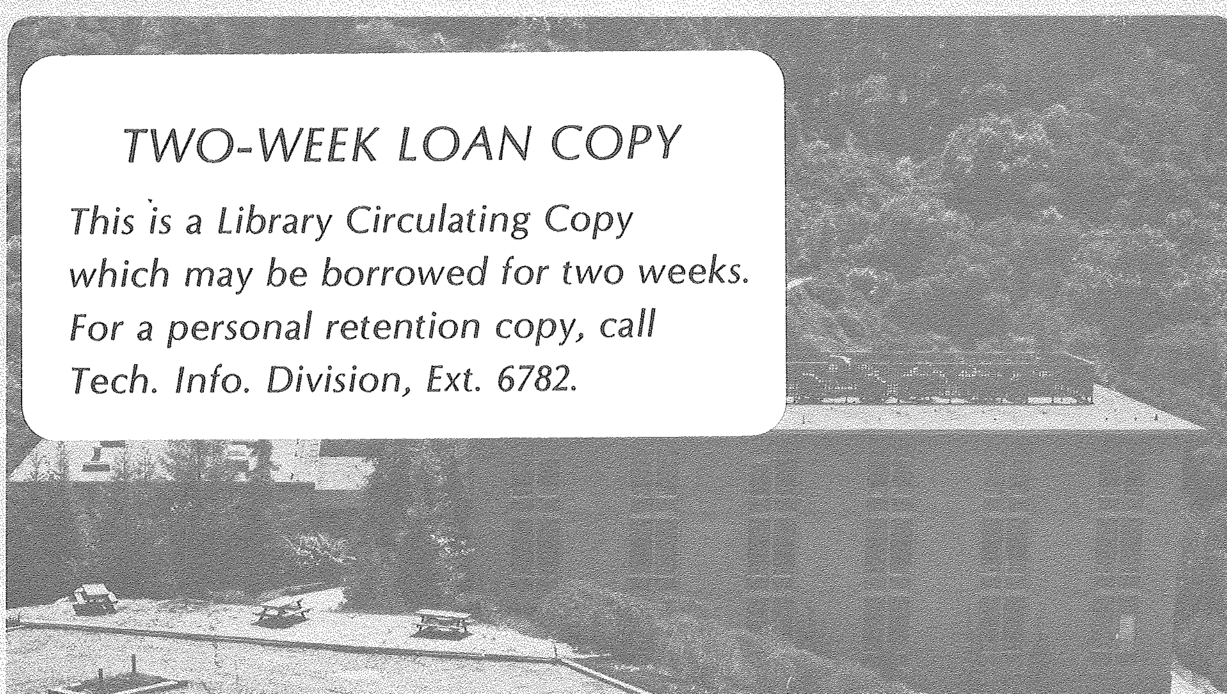

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EFFECTS OF MORPHOLOGY ON THE MECHANICAL BEHAVIOR OF DUAL PHASE Fe/Si/C STEELS

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ABSTRACT

A study has been made on the effects of morphology on the mechanical behavior of dual phase Fe/2Si/0.1C steels. The coarse dual phase structure obtained by continuously annealing in the two phase region directly from the austenite region results in poor elongation ductility with relatively high strength. However, upon obtaining a fine fibrous or fine globular dual phase structure by following different transformation paths, significant improvements occur in elongation ductility without much sacrifice in strength. The poor elongation ductility of the coarse dual phase structure is due to the initiation of cleavage cracks in the ferrite region where maximum localized stress concentration took place. But, in steels with both fine fibrous or globular morphologies, fracture occurred by void nucleation and coalescence after large amounts of plastic deformation.

I. INTRODUCTION

Dual phase steels whose structures consist of ferrite and martensite have received great attention in the past few years because of their characteristic mechanical properties. It has been shown that at a given tensile strength level, these dual phase steels have superior formability to commercially available high strength low alloy steels.[1-9]

Several important points regarding the deformation mechanisms in dual phase steels have already been documented. The continuous yielding has been attributed to the presence of mobile dislocations in the ferrite matrix.[1,2,3] These dislocations are by-products of the austenite to martensite transformation. Additionally, tensile strength has been shown to vary linearly with martensite volume fraction by the mixture rule[1,2,4] with the exception of steels having fine precipitates in ferrite.[4,5] Also, the total elongation has been found to depend on the martensite volume fraction.[3,4] From the viewpoint that the mechanical properties, in general, are sensitive to the microstructures of the system, studies have been made to correlate the microstructure with the mechanical properties of dual phase steels.[6,7] However, the role of morphology (size, shape, distribution) of the constituents in dual phase steels is not yet clear. Therefore, the aim of this investigation is to clarify the effects of the morphology on the mechanical behavior of dual phase steels.

II. EXPERIMENTAL PROCEDURE

The material used in this investigation was high purity Fe/2Si/0.1C steel. This steel was homogenized under an argon atmosphere at 1100°C for 24 hours, and then furnace-cooled. Tensile and transmission electron microscopy specimens were prepared by cutting the sheets into slightly bigger dimensions than the exact tensile specimens, and were heat treated as described later. Transmission electron microscopy specimens made by twin-jet polishing were examined in a Philips EM301 microscope and also in a Philips EM400 microscope. The latter was used for X-ray analysis of the silicon distribution in the ferrite and martensite.

Fracture surfaces of tested specimens were examined with an AMR 1000 scanning electron microscope. Also, to clarify where cracks are first nucleated and propagated, tensile specimens at the various stages of deformation were cut along the tensile axis and examined normal to the tensile axis with an AMR 1000 scanning electron microscope. All tensile tests were performed using 1.25" gage round specimens at room temperature on an Instron with a cross head speed of 0.05cm/min.

In order to obtain dual phase steels with various morphologies, three kinds of heat treatments were used as shown in Fig. 1. For the intermediate quenching treatment, samples were held for 1 hr. at 1100°C and subsequently quenched into iced brine while for the step quenching treatment, samples were heat treated directly into the two phase ($\alpha+\gamma$) region after the austenitizing treatment. The two phase annealing temperatures were chosen so as to obtain the required volume fraction of martensite. Samples were held at the chosen two phase annealing temperature for twenty minutes and iced brine quenched.

III. RESULTS

1. Metallography

A. Optical microstructure

As expected from the fact that the microstructures of dual phase steels were mainly determined by their transformation paths, there are remarkable differences in the resulting morphologies which are obtained by various heat treatments. These are clearly shown in Fig. 2. In the intermediate quenching process (Fig. 1a), reheating to the two phase range after quenching from austenite results in the nucleation of austenite along the lath boundaries of prior martensite. This results in a distribution of fine fibrous martensite in the ferrite matrix as shown in Fig. 2a. In the intercritical annealing process (Fig. 1b), the initial microstructure is hypoeutectoid ferrite and pearlite. Upon two phase annealing, austenite nucleates at the carbide/ferrite interfaces and grows,[10] giving rise to the distribution of fine globular martensite along the ferrite boundaries in the resulting structure as shown in Fig. 2b. In the step quenching process (Fig. 1c), the phase prior to two phase annealing is austenite. Upon decreasing the temperature to the two phase range, ferrite nucleates at the prior austenite grain boundaries and grows into the austenite. The resulting structure is such that coarse martensite is surrounded by ferrite as shown in Fig. 2c. The microstructure thus depends on the prior austenite grain size which is large and thus tends to be coarser than those obtained from martenstitic or hypoeutectoid structures.

B. Transmission Electron Microscopy

Careful transmission electron microscopy studies on each structure show no remarkable difference in their structures. Fig. 3 shows typical dislocated martensite with interlath retained austenite. Also, microtwins were often found in some areas. These particular micrographs were obtained form the step quenched coarse dual phase structure, but the substructure of the martensite was similar in all cases. The transmission electron micrographs of the ferrite region show the formation of subgrains in the intermediate quenched fine fibrous structure and in the intercritically annealed fine globular structure, while in the step quenched coarse dual phase structure, it is not found. This is shown in Fig. 4.

2. Mechanical Properties

The strength and the elongation values as a function of martensite volume fraction for various dual phase structures are shown in Fig. 5 and Fig. 6, respectively. At the same volume fraction of martensite, the coarse dual phase structure (Fig. 2c) obtained by step quenching results in very poor elongation ductility with relatively high strength value as compared to the other microstructures (Fig. 2a, b).

Scanning electron micrographs of fracture surfaces are shown in Fig. 7. Coarse dual phase structures fracture predominantly by cleavage as revealed by well-defined facets, while both fine fibrous and fine globular structures fracture in a ductile manner. Thus, it is clear that the mechanical properties of the dual phase steels are largely dependent upon their morphologies.

IV. DISCUSSION

There are many factors which affect the mechanical properties of Fe/Si/C and dual phase steels in general, e.g., 1) morphology of ferrite and martensite, 2) properties of ferrite and martensite, 3) volume fraction of martensite. The transformation paths control these main structural features.

It is well known that the tensile strength of dual phase steel is linearly dependent upon the martensite volume fraction by the mixture rule. Even if this mixture rule describes the extreme case where the strong phase particles are continuous and unidirectionally aligned in the direction of the applied stress,[11] it shows good agreement with the experimental values. However, it has been shown that the slope of strength variation with changing martensite volume fraction is different with different alloy systems and with different microstructures.[4,5] This behavior has also been confirmed in the present work (Fig. 5).

The analysis of the solute distribution in the ferrite and martensite of various dual phase structures by STEM microanalysis has shown that silicon was uniformly distributed in the ferrite and martensite in every case. This was expected because of the low diffusivity of silicon at the two phase annealing temperature. From the lever rule in the phase diagram, carbon distribution in the various structures is expected to be the same at the same volume fraction of martensite. Also, as seen in transmisssion electron micrographs, there is no significant difference in the substructures after the various dual phase treatements. Thus, it seems that this behavior is solely due to the different morphologies obtained after the various transformation paths. As shown in Fig. 6, coarse dual phase structures show very low elongation values within the range of martensite volume fraction investigated. These low elongation values of coarse dual phase structures are found to be due to cleavage crack initiation and propagation in the ferrite matrix which in turn causes premature failure of the system at the early stage of deformation as shown in Fig. 8. This is quite surprising because it has been generally believed that failure of dual phase steels is initiated by fracture in the martensite region. Karlsson and Sundström[12] showed that the maximum strain develops inside the ferrite matrix and is often located away from the martensite-ferrite interfaces. The flow strength of the ferrite matrix is much lower than that of the martensite, so that plastic deformation begins in the soft ferrite while the martensite is still elastic. This plastic deformation in the ferrite is constrained by the adjacent martensite, giving rise to tangles of dislocations at the martensite-ferrite interfaces and hence a build-up of stress concentration in the ferrite. Thus, these localized deformations and/or stress concentrations in the ferrite lead to fracture of the ferrite matrix which occurs by cleavage or void nucleation and coalescence depending on their morphological differences. The latter type of soft phase fracture in two phase systems has also been observed in non-ferrous alloy systems.[13,14] As shown in Fig. 7a, coarse dual phase structures fracture by cleavage indicating that fracture of ferrite was due to the highly localized stress concentration in the ferrite with small amounts of plastic deformation. Concurrent slip line studies confirm that plastic flow occurs in the ferrite, but is rather inhomogeneous. However, at the same volume fraction of martensite, either fine fibrous or fine globular strcutures have smaller inter-second phase spacings which, in turn, decrease the density of dislocations and also the stress concentration which in turn decreases the probability of cleavage crack nucleation in the ferrite matrix. Instead, fracture of ferrite in these structures occurs by void nucleation and coalescence due to localized deformation in the ferrite after large amounts of plastic deformation. In the fine fibrous structure, a large density of void nucleation was found near the martensite-ferrite interface and voids did not necessarily nucleate because of fracture of martensite (Fig. 9). Fig. 10 shows the sequence of crack propagation in this structure. These micrographs were obtained by successively polishing and examining the surface. In Fig. 10a, the crack is propagating along the martensite-ferrite interface (see arrow), and below that surface the crack is propagating through the martensite (Fig. 10b). However, by further polishing the surface, it was found that the crack stops its propagation at the martensite-ferrite interface (Fig. 10c). These micrographs suggest that the crack propagated into the martensite after it had propagated in the ferrite. The same fracture mechanism operated in the fine globular dual phase structure. Fig. 11a shows void formation near the martensite-ferrite interfaces. Such cracks, formed along the interface, propagated within the ferrite matrix with increasing deformation (Fig. 11b). This characteristic process of crack propagation seems to be due to the properties of martensite, which is plastically deformable, whereas in pearlitic and similar structures in plain carbon steels fracture is determined by the cementite. This phase is not present in these dual phase steels. The considerable difference in the fracture behavior between the coarse and the fine dual phase structures can explain a marked difference in the ductility of the bulk materials at room temperature. However, the cause for the higher strength of coarse dual phase structures is not yet clear. Also, it seems that the shape (fibrous, globular) of the martensite does not have a marked influence on the mechanical behavior of dual phase steels. The foregoing discussion emphasizes that the properties of dual phase steels are largely dependent upon their morphologies, and, especially grain size.

V. SUMMARY

It has been found that the ferrite-martensite morphology has a great effect on the mechanical behavior of dual phase steels.

1) The coarse dual phase structure has a much lower elongation ductility, but higher strength than the fine fibrous and fine globular

dual phase structures.

2) The failure of coarse dual phase structure is due to the initiation of cleavage cracks in the ferrite matrix, while in the fine fibrous and fine globular structures, failure occurs by void nucleation and coalescence.

3) In all cases, cracks propagate primarily in the ferrite matrix rather than into the martensite.

4) The importance of fine grains should be again emphasized in producing the best combinations of strength and ductility in dual phase steels. However, it seems that the shape of martensite has little effect on mechanical behavior of dual phase steels.

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FIGURE CAPTIONS

Fig. 1 Schematic representation of dual phase heat treatments

a) Intermediate quenching b) Intercritical annealing c) Step quenching

Fig. 2 Optical micrographs of dual phase structures. XBL 801 4610

a) Intermediate quenching b) Intercritical annealing c) Step quenching XBB 801 1373

Fig. 3 Transmission electron micrographs showing typical dislocated martensite with interlath retained austenite, which shows K-S orientation relationship, $<111>_{\alpha}//<110>_{\gamma}$. These particular micrographs were obtained from a step quenched specimen. XBB 801 1375

Fig. 4 Dislocation substructures in ferrite region after

a) Intermediate quenching b) Intercritical annealing XBB 801 1374

Fig. 5 Yield and ultimate tensile strengths as a function of martensite volume fraction. XBL 801 4608

Fig. 6 Uniform and total elongations as a function of martensite volume fraction. XBL 801 4609

Fig. 7 Scanning electron fractographs of fractured tensile specimen

a) Intermediate quenching b) Intercritical annealing

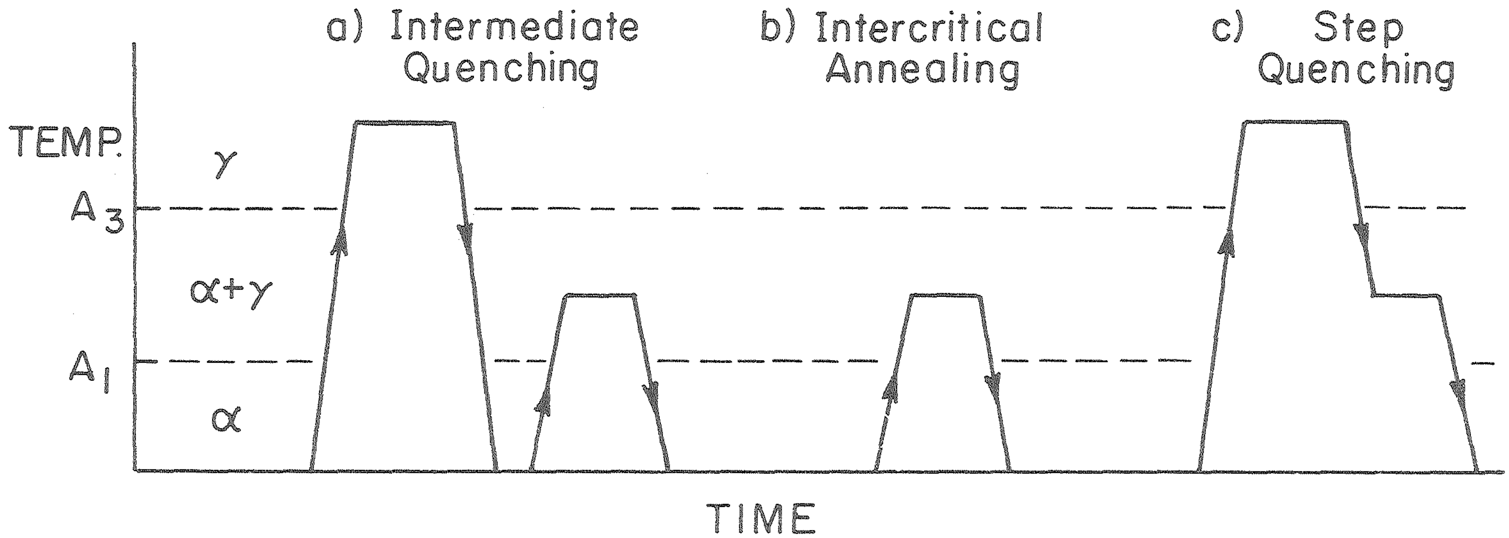
c) Step quenching XBB 801 1372

Fig. 8 Scanning electron micrograph of step quenched coarse dual phase structure showing the cleavage cracks developed in ferrite M: martensite, F: ferrite. The stress axis is normal to the micrograph. XBB 804 4090

Fig. 9 Scanning electron micrograph of intermediate quenched fine fibrous structure showing void formation near the martensite-ferrite interface. The stress axis is normal to the micrograph. XBB 804 4091

Fig. 10 Scanning electron micrographs of intermediate quenched fine fibrous structure showing the sequence of crack propagation. The stress axis is normal to these micrographs. XBB 802 19841

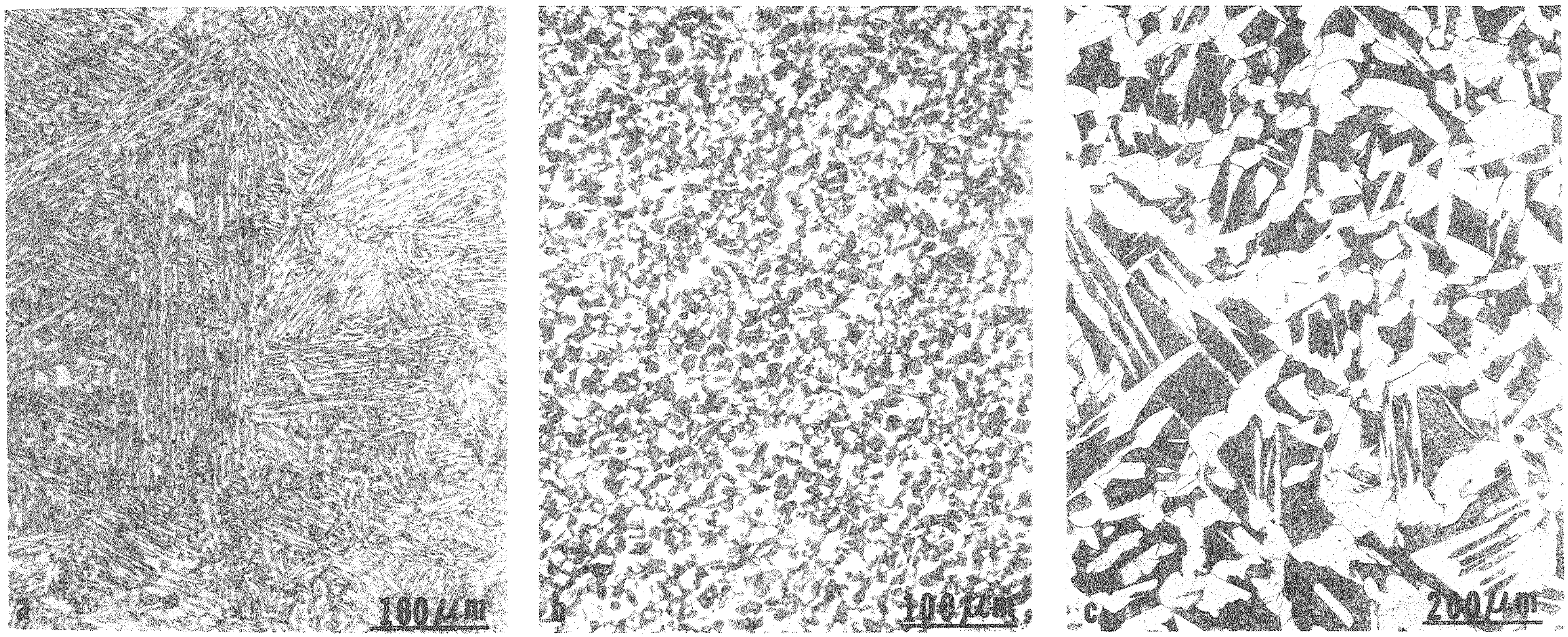
Fig. 11 Scanning electron micrographs of intercritically annealed fine globular structure showing void formation near the martensite-ferrite interface (a) and crack propagation in the ferrite matrix (b). The stress axis is normal to the micrographs indicating shear fracture rather than cleavage. XBB 804 4089



XBL 801-4610

Fig. 1
(a,b,c)

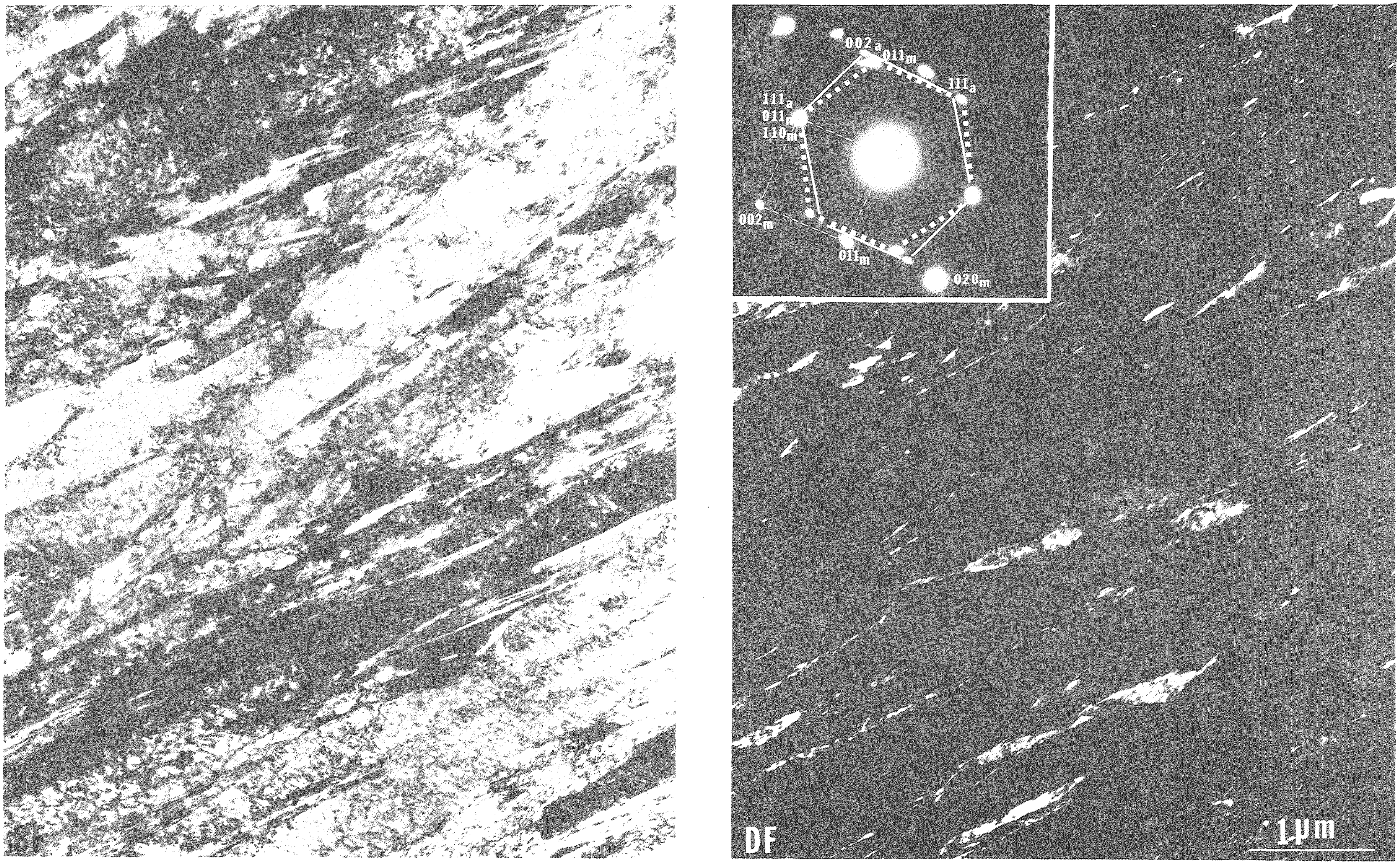
OPTICAL MICROGRAPH



INTERMEDIATE QUENCHING   INTERCRITICAL ANNEALING   STEP QUENCHING

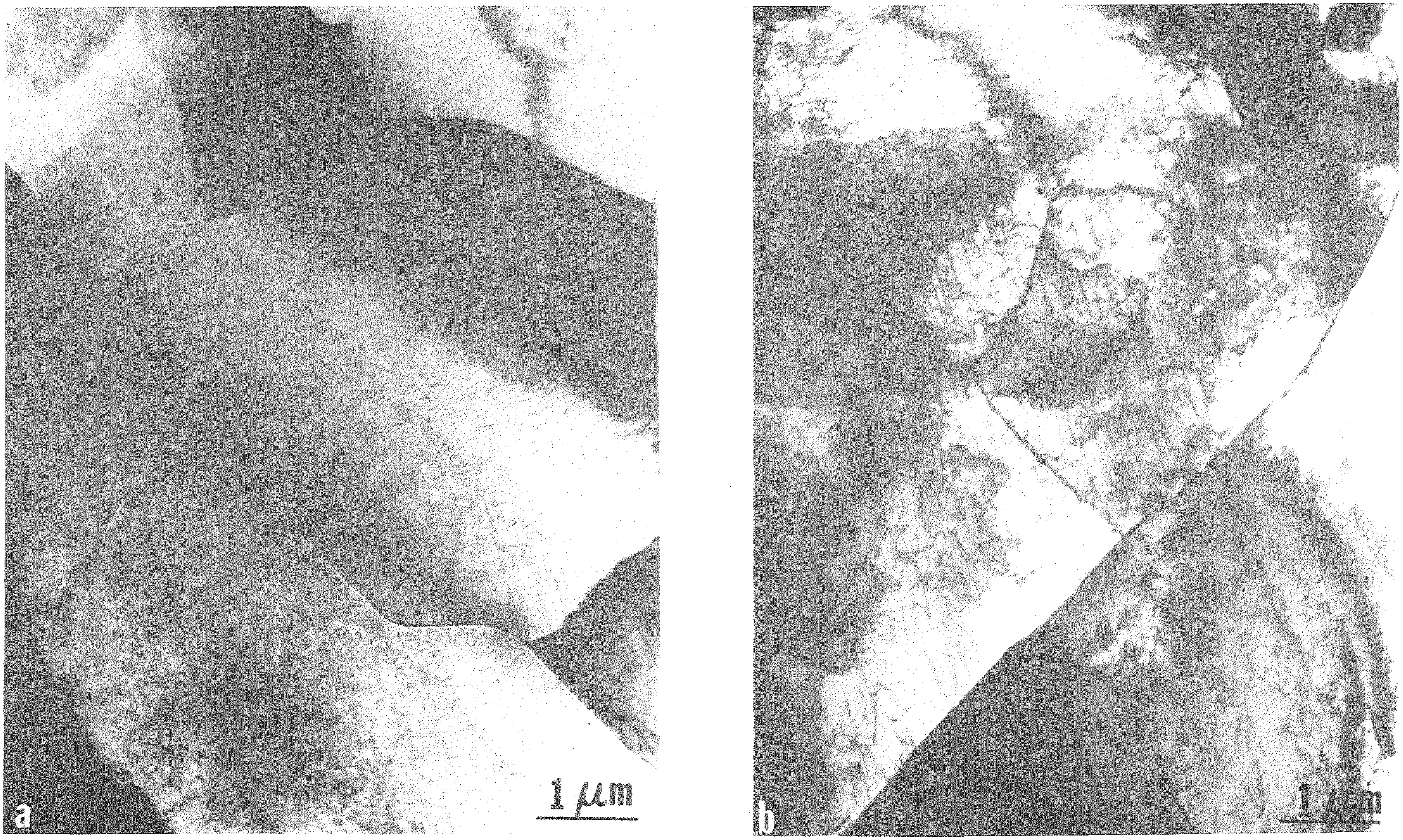
XBB 801-1373

Fig. 2
(a,b,c)



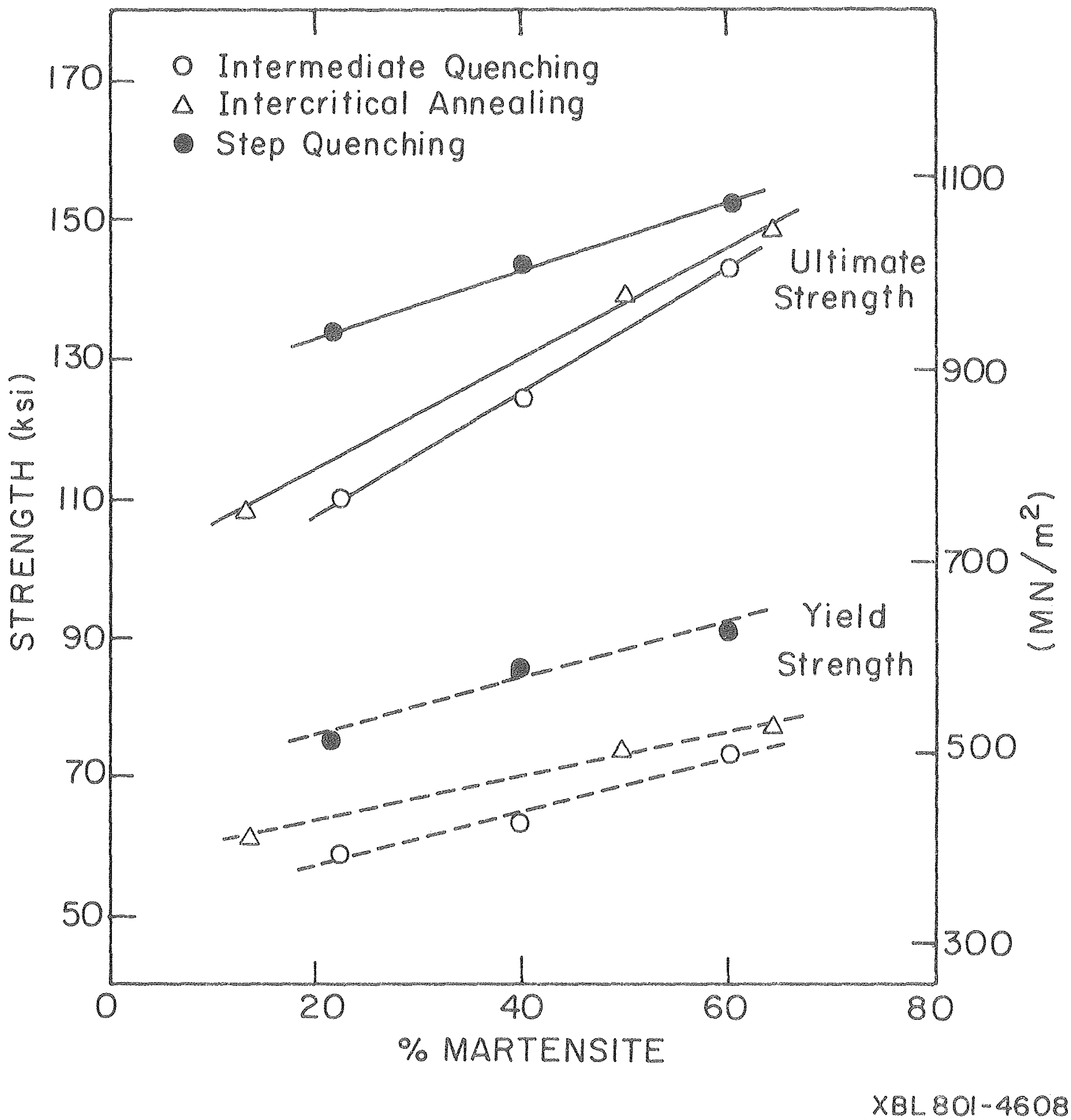
XBB 801-1375

Fig. 3
(a,b)



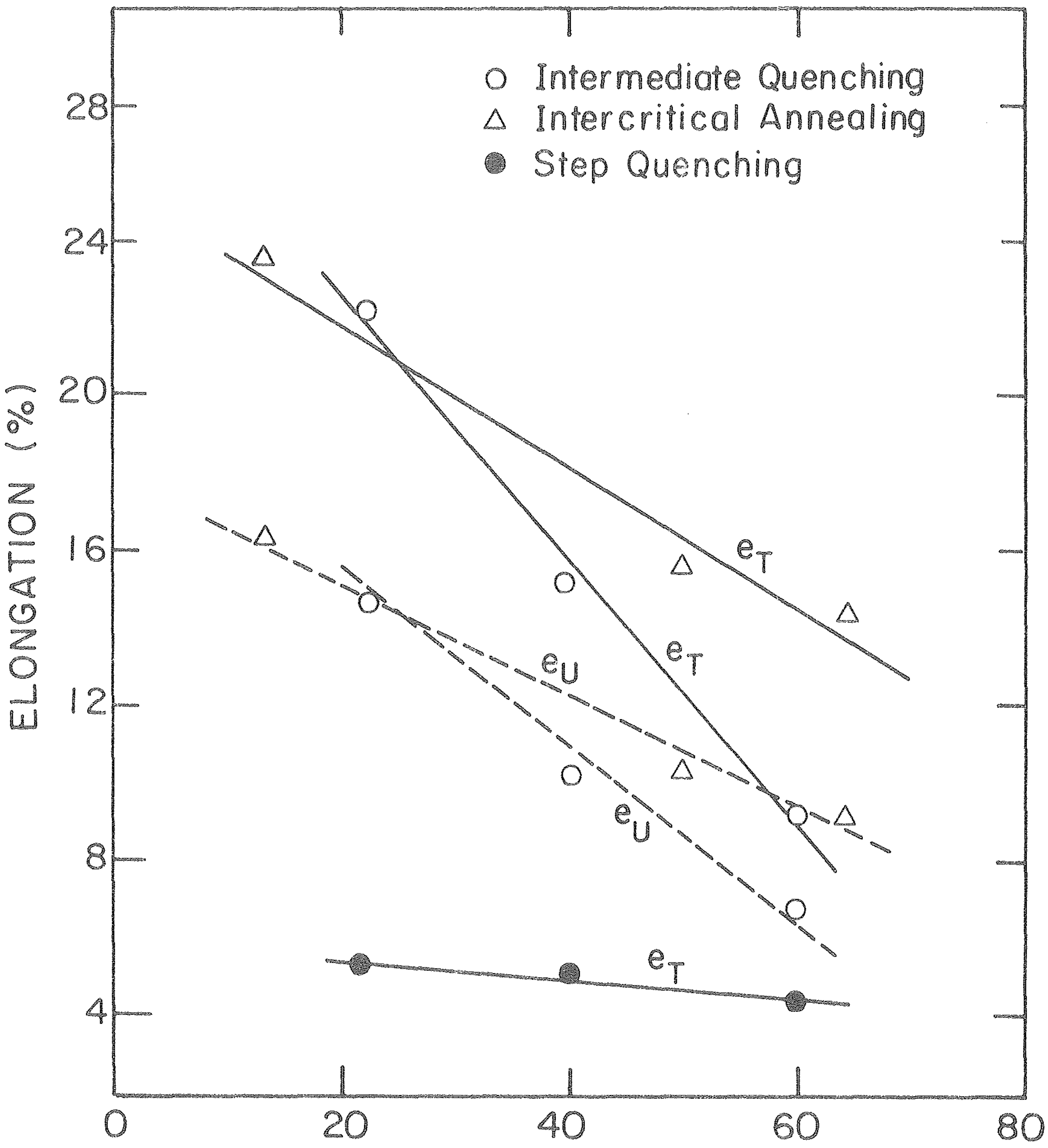
XBB 801-1374

Fig. 4
(a & b)



Intermediate Quenching
Intercritical Annealing
Step Quenching
STRENGTH (ksi)
170
150
130
110
90
70
50
1100
900
700
500
300
(MN/m²)
Ultimate Strength
Yield Strength
0
20
40
60
80
% MARTENSITE
XBL 801-4608

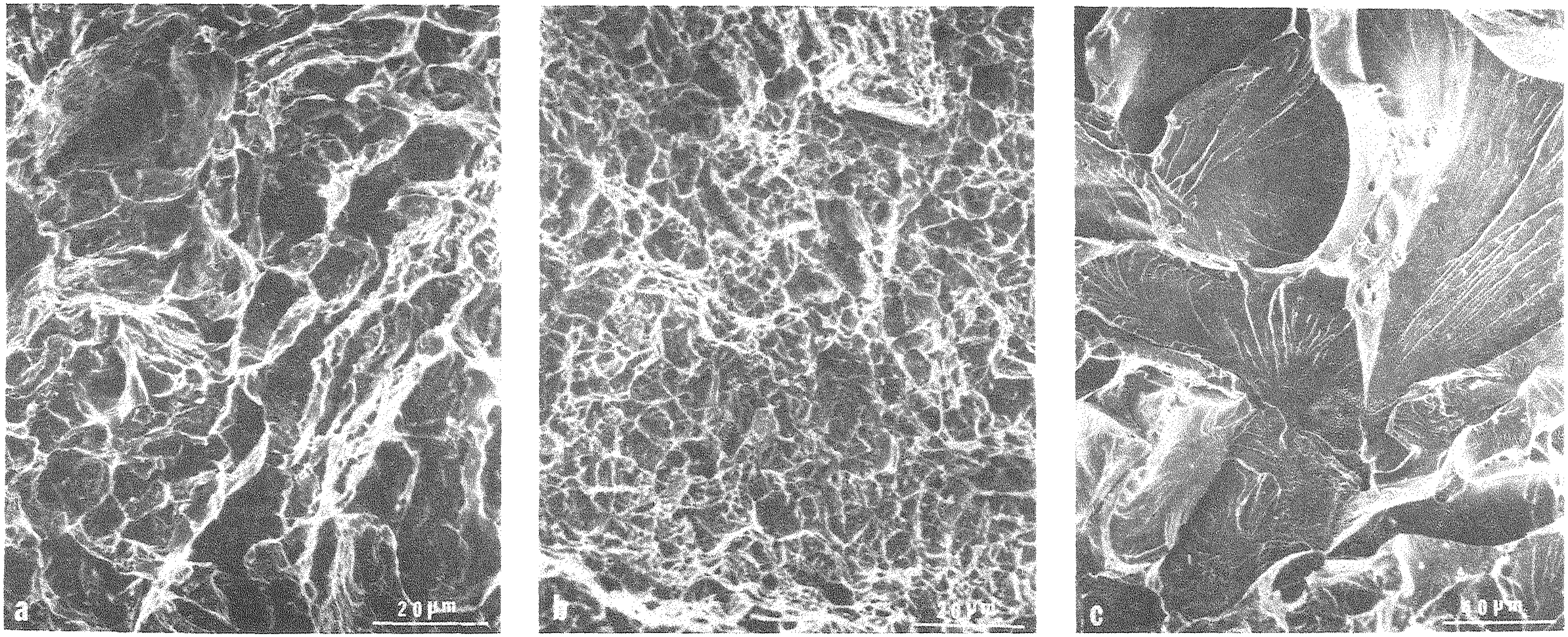
Fig. 5



XBL 801-4609

Fig. 6

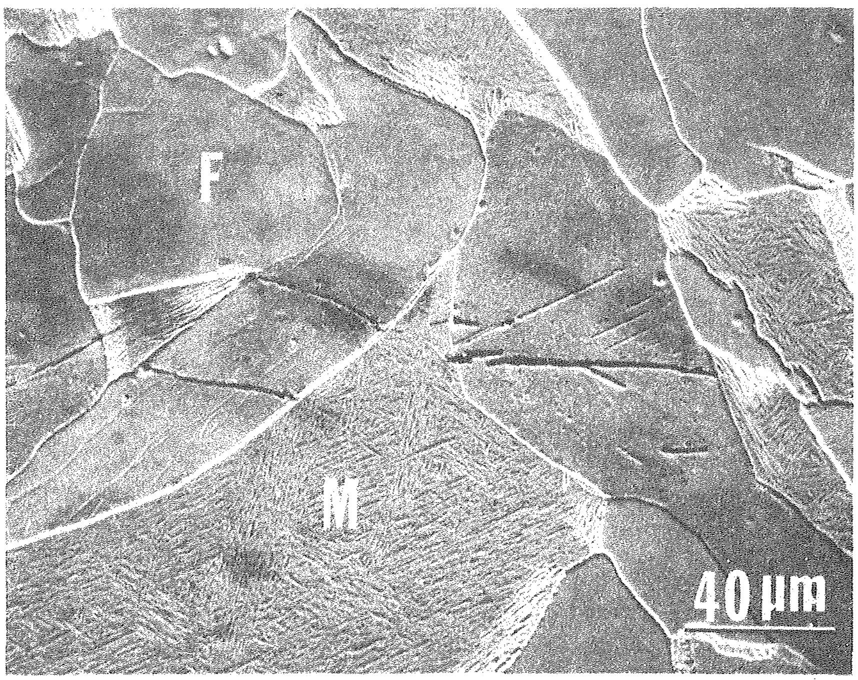
SEM FRACTOGRAPH



INTERMEDIATE QUENCHING | INTERCRITICAL ANNEALING | STEP QUENCHING

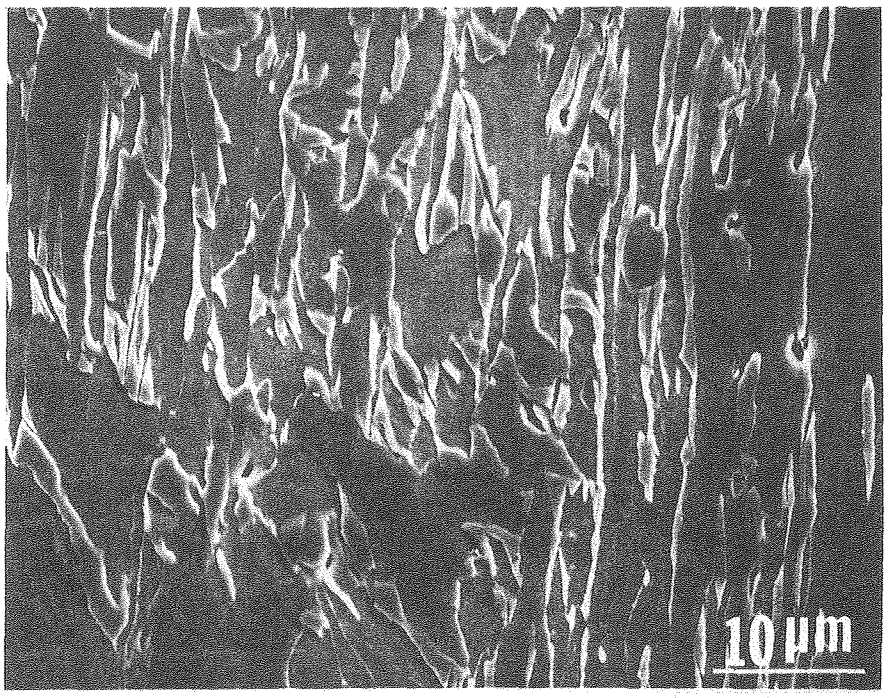
XBB 801-1372

Fig. 7
(a,b,c)



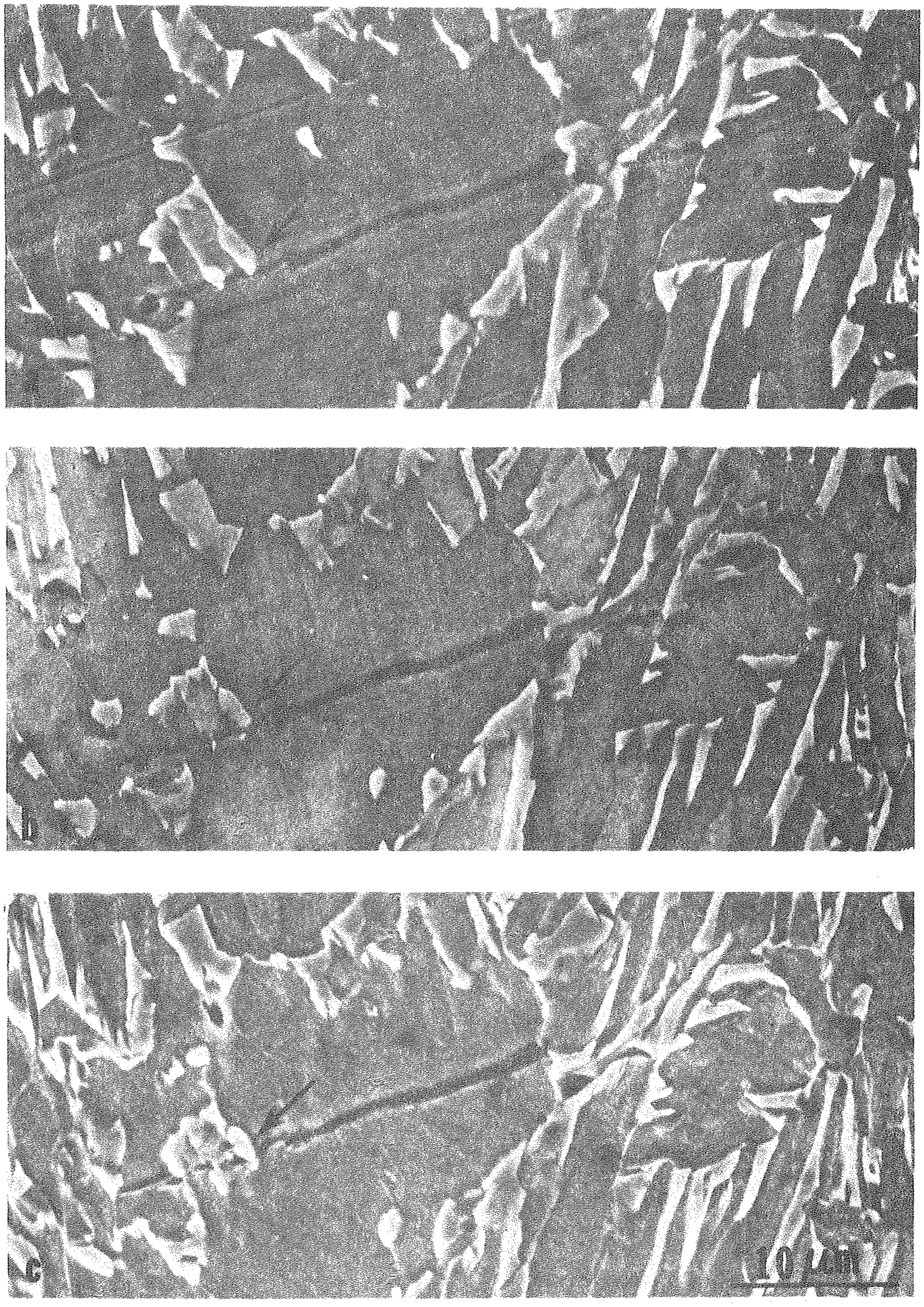
XBB 804-4090

Fig. 8



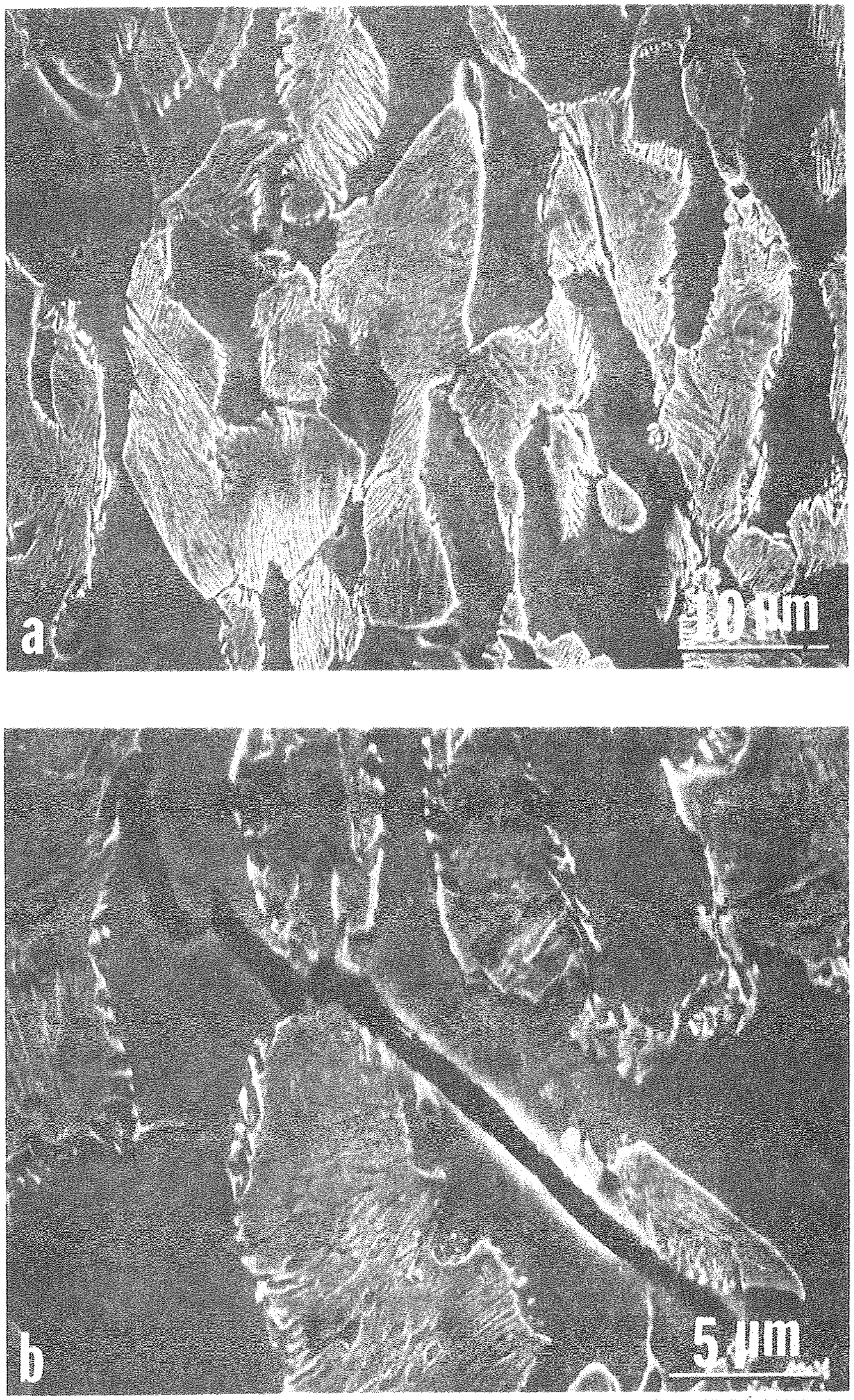
XBB 804-4091

Fig. 9



XBB 802-1984

Fig. 10
(a,b,c)



XBB 804-4089

Fig. 11
(a & b)