Title

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DISCUSSION: "TOUGHNESS VARIATIONS DURING THE TEMPERING OF A PLAIN CARBON MARTENSITIC STEEL" BY J. E. KING, R. F. SMITH AND J. F. KNOTT*

RETAINED AUSTENITE AND TRANSGRANULAR TEMPERED MARTENSITE EMBRITTLEMENT

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The following discussion is presented to substantiate, clarify and reinterpret some of the results of the above authors for the existence and origin of tempered martensite embrittlement (TME) in relation to retained austenite, with reference to the experimental steels we have been working for the past several years[1]. McMahon and Thomas[2] first identified, through very careful electron metallographic analysis, stabilized films of retained austenite, and showed that the decomposition of this retained austenite on tempering caused a drop in toughness. In this report, we summarise some of our latest data on this effect. All the alloys investigated were vacuum melted using high purity raw materials to minimize the trace impurity content (individual percentages of S and P were <0.01 w/o in all the alloys).

A monotonic increase in the volume fraction of retained austenite in the as-quenched structures (refer to references 1 through 3 for experimental details) as a result of quaternary alloy additions of Mn up to 2 w/o to the ternary Fe-4Cr-0.3C alloy was found. The electron micrographs (e.g., Fig. 1) reveal typical continuous interlath films of retained austenite in all the alloys. This austenite in the Mn

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modified alloys was found to be stable to transformation up to tempering temperatures below 300°C. Following tempering at 300°C, however, it transforms to coarse interlath carbides as shown in Fig. 2. The accompanying variations in toughness properties are presented in Fig. 3. It is clear that the % reduction in area, a measure of tensile ductility, shows two minima, viz., one at 300°C (transgranular fracture) and the other at 500°C (intergranular fracture), for the 2%Mn modified alloy, whereas no indication of a minimum in this property is discernible for the 0.5%Mn modified alloy [Fig. 3(a)]. However, the Charpy impact toughness [Fig. 3(b)] clearly reveals a toughness minimum at 300°C tempering for the 0.5% Mn alloy while for the 2%Mn alloy the minimum occurs only at 500°C although severe toughness degradation occurs in this alloy following 300°C tempering. Fractography (Fig. 4) revealed that the predominantly dimpled rupture in both the alloys at 200°C [Fig. 4(a)] is quickly replaced by transgranular quasi-cleavage fracture in the 300°C tempered specimens [Fig. 4(b)]. The quasi-cleavage fracture at 300°C [Fig. 4(b)] indeed reveals some parallel ridges indicative of a fracture path along interlath boundaries which are weakened by the precipitation of coarse carbides (Fig. 2). The 2%Mn alloy showed increasingly brittle fracture for tempering above 300°C and at 500°C, the fracture was completely intergranular. This behavior is probably due to segregation effects that lower the intercrystalline cohesion which is one of the widely proposed mechanisms for temper embrittlement.

The present results show that the nature of toughness measurements in terms of strain rate and local stress conditions can be important to

be able to detect any embrittlement minimum in that property. The 2%Mn alloy had a higher percentage of retained austenite and, therefore, would produce a higher volume fraction of interlath coarse carbides following its transformation. Thus, a more severe embrittlement is expected in the 2%Mn alloy compared to 0.5%Mn alloy which is substantiated by the occurrence of toughness minima in both the % reduction in area as well as the impact toughness for the former, but only a toughness minimum occurs in the impact toughness for the latter. The strain rate in a tensile (or K_{Ic}) test is about a million times smaller than that in an impact test and a higher strain rate is known to increase the DBTT. At very low strain rates (such as in a tensile test), a toughness minimum may not be detected unless the embrittlement is very severe [compare Figs. 3(a) and (b)].

In our research we have also found that both TME (transgranular cleavage fracture) and temper embrittlement (intergranular fracture) can occur in the same alloy depending on its composition (e.g., the 2%Mn modified alloy). On the other hand no TME was observed in Fe/Mo/C steels in which retained austenite was not detected[3]. Thus TME appears to be a more general phenomenon[1] in alloys having appreciable amounts of retained austenite while temper embrittlement is specific to a few alloy compositions. A detailed microstructural characterization is vital for pinning down the origin of TME. In the investigation under discussion, the authors chose a 0.6%C alloy. Although King et al. did not report any transmission electron microscopy, others[4] have shown that the morphology is mixed viz., about 25-30% plate martensite, 60% lath, and close to 8% retained austenite. The evidence for substantial amounts of plate martensite in the structure can also be derived from the above authors' observation of microcracks. The morphology

of retained austenite occurring in plate martensite is quite unlike that in the lath martensites found in our alloys. The expected differences are illustrated in the schematic diagram of Fig. 5. The island morphology of retained austenite [Fig. 5(a)] in plate martensite is expected to give rise to a carbide morphology and distribution shown in Fig. 5(b) upon tempering. This is quite different from that occurring in lath martensite [Figs. 2,5(d)] following austenite decomposition. In the former, the islands containing coarse carbide precipitates should provide easy crack paths and thus, the fracture surface should contain islands of cleavage fracture which indeed appears to be the case from the published fractograph, Fig. 4, of the above paper.

In summary, tempered martensite embrittlement in the range 300-400°C is associated with decomposition of retained austenite. The fracture is transgranular with respect to prior austenite but is clearly intergranular with respect to the individual martensite laths.

Acknowledgement

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References

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2. J. A. McMahon and G. Thomas, "The Microstructure and Design of Alloys", Proc. of the 3rd Int'l. Conf. on the Strength of Metals and Alloys, 1 , p. 180, Inst. of Metals, Cambridge, England (1973).

3. R. A. Clark and G. Thomas, Metallurgical Trans., 6a (1975), 969-979.

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Figure Captions

Fig. 1. (a) Bright-field and (b) dark-field micrographs showing retained austenite in as-quenched Fe-4Cr-0.3C-2Mn alloy.

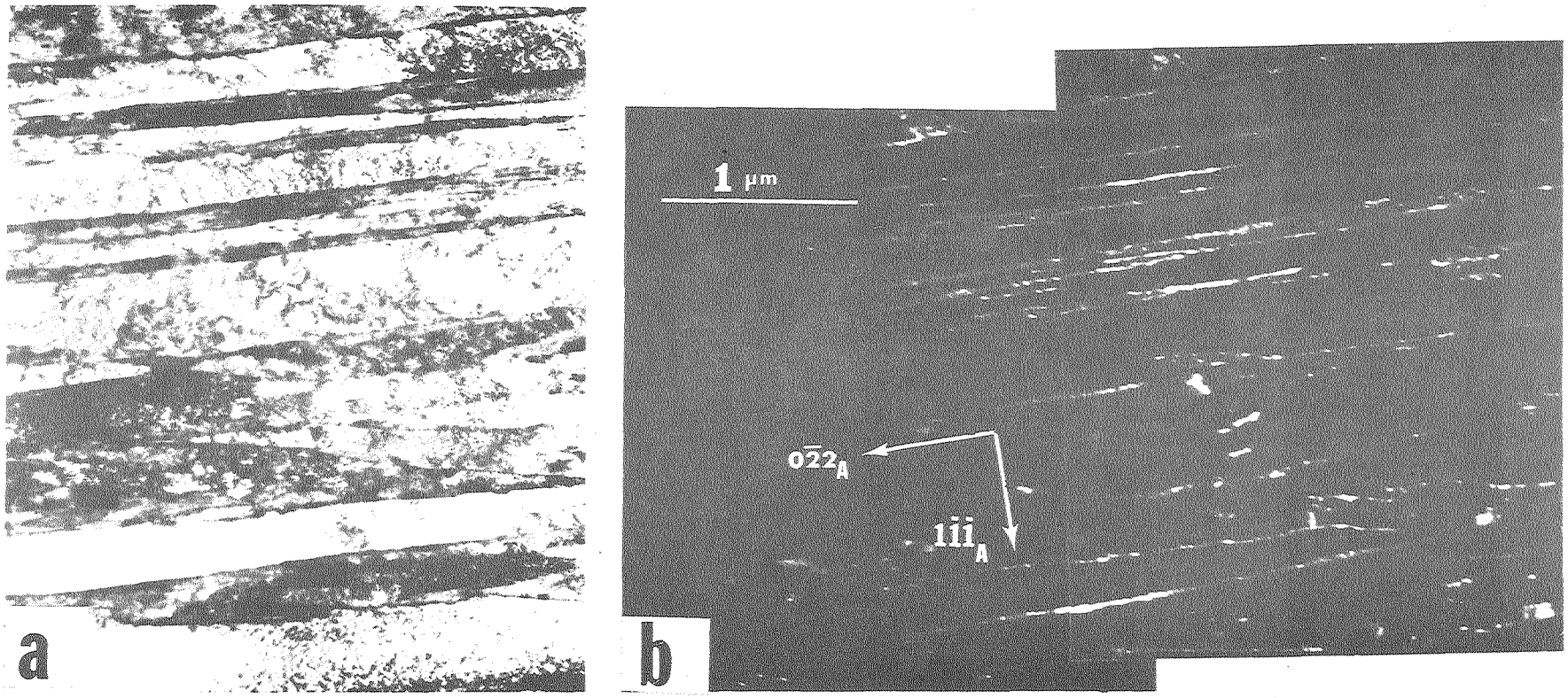
Fig. 2. (a) Bright-field and (b) dark-field micrographs revealing both interlath (marked by arrows) as well as intra-lath carbides (M_3C) in Fe-4Cr-0.3C-0.5Mn alloy tempered 1 hr. at 300°C.

Fig. 3. (a) % reduction in area vs. tempering temperature. The retained austenite percentages are indicated on the curves.

(b) Charpy-V-Notch impact toughness vs. tempering temperature. A,B,C,D indicate points where significant changes in toughness occur.

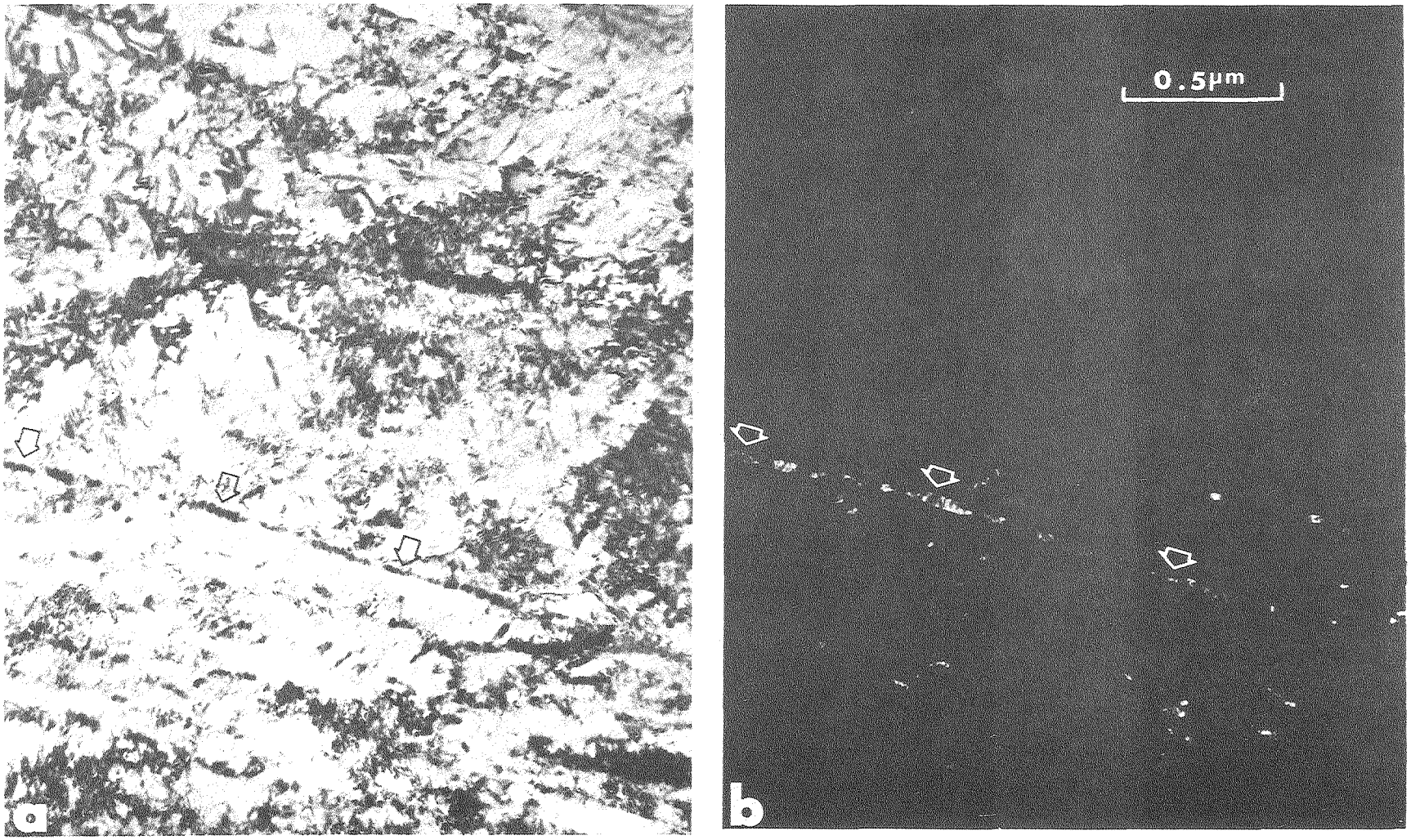
Fig. 4. Fractographs (Charpy) showing dimpled rupture for 200°C tempering (a) and transgranular quasi-cleavage (b) for 300°C tempering in the Fe-4Cr-0.3C-2Mn alloy.

Fig. 5. Schematic illustrations of the morphology of retained austenite and martensite in high carbon, high alloy steels (a) and low carbon, low alloy steels (c). Morphology of fracture due to decomposition of retained austenite → carbide on tempering [(b) and (d)] depends on the initial microstructure (see text).



XBB 775-4291

Fig. 1



XBB 776-5738

Fig. 2

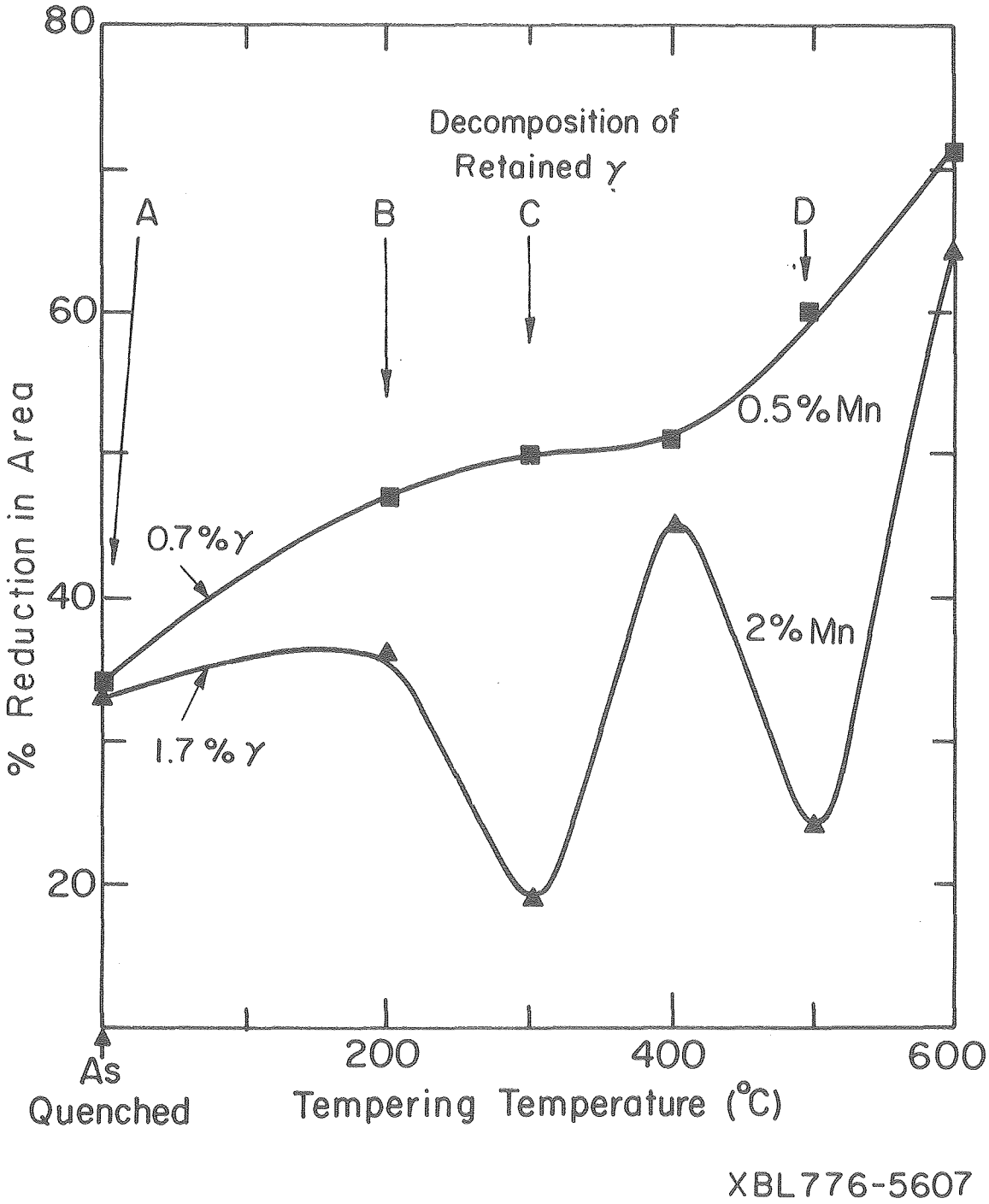

80
60
40
20
% Reduction in Area
Decomposition of
Retained γ
A
B
C
D
0.5% Mn
0.7% γ
2% Mn
1.7% γ
As
Quenched
200
400
600
Tempering Temperature (°C)
XBL776-5607

Fig. 3a

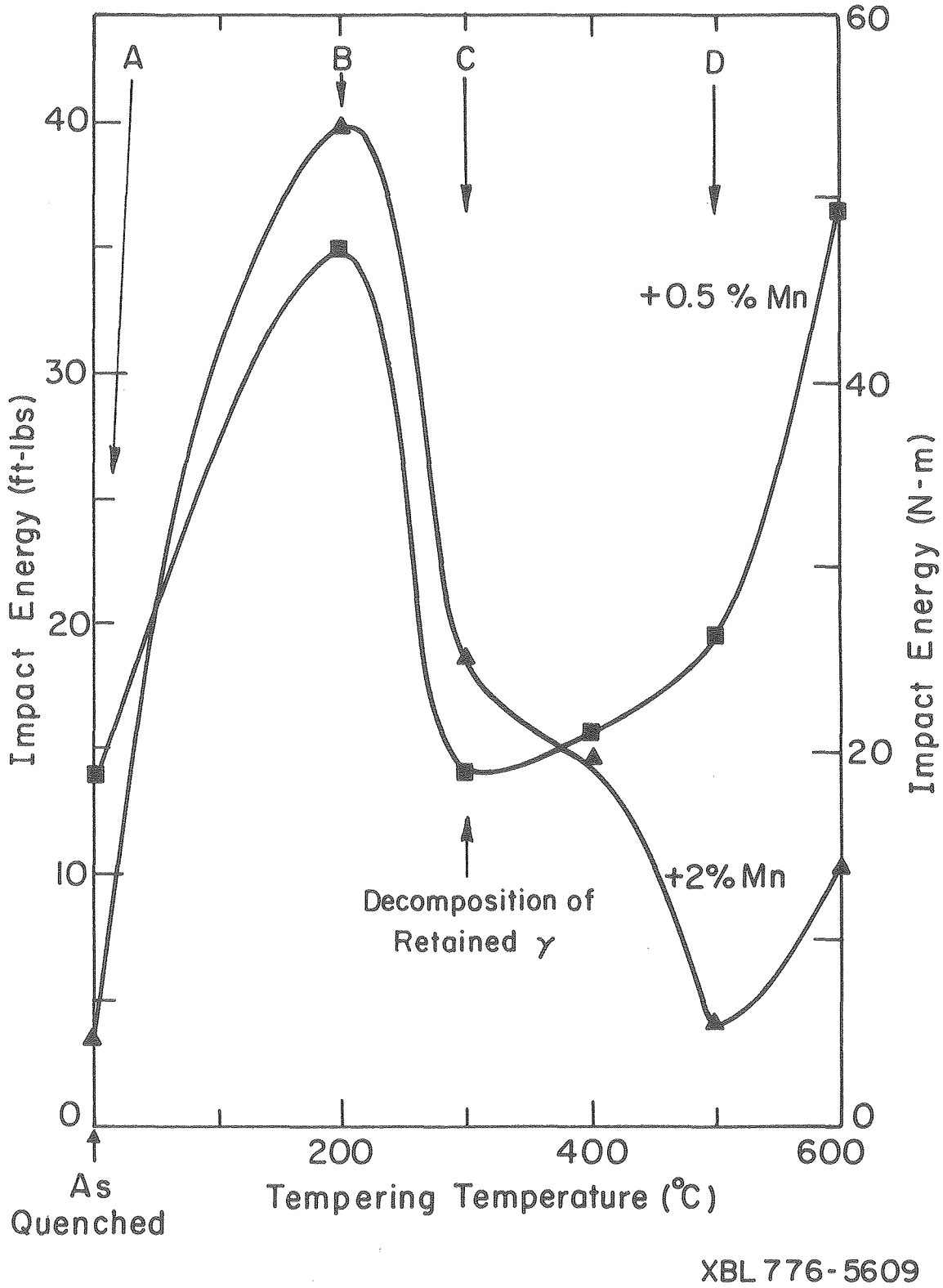
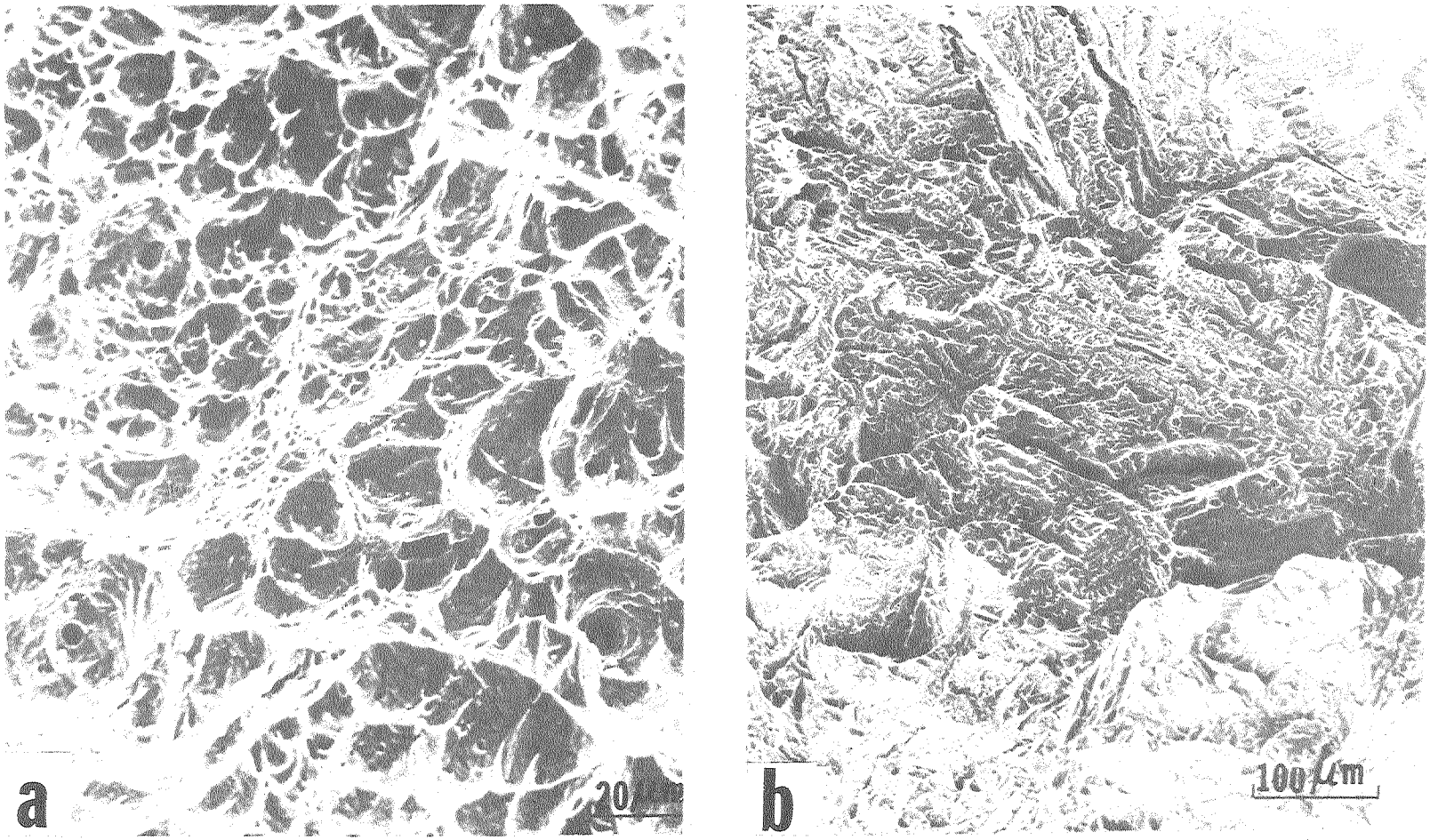

A
B
C
D
+0.5 % Mn
+2% Mn
Decomposition of
Retained γ
Impact Energy (ft-lbs)
Impact Energy (N-m)
0
10
20
30
40
0
20
40
60
200
400
600
As
Quenched
Tempering Temperature (°C)
XBL 776-5609

Fig. 3b



XBB 774-3110

Fig. 4

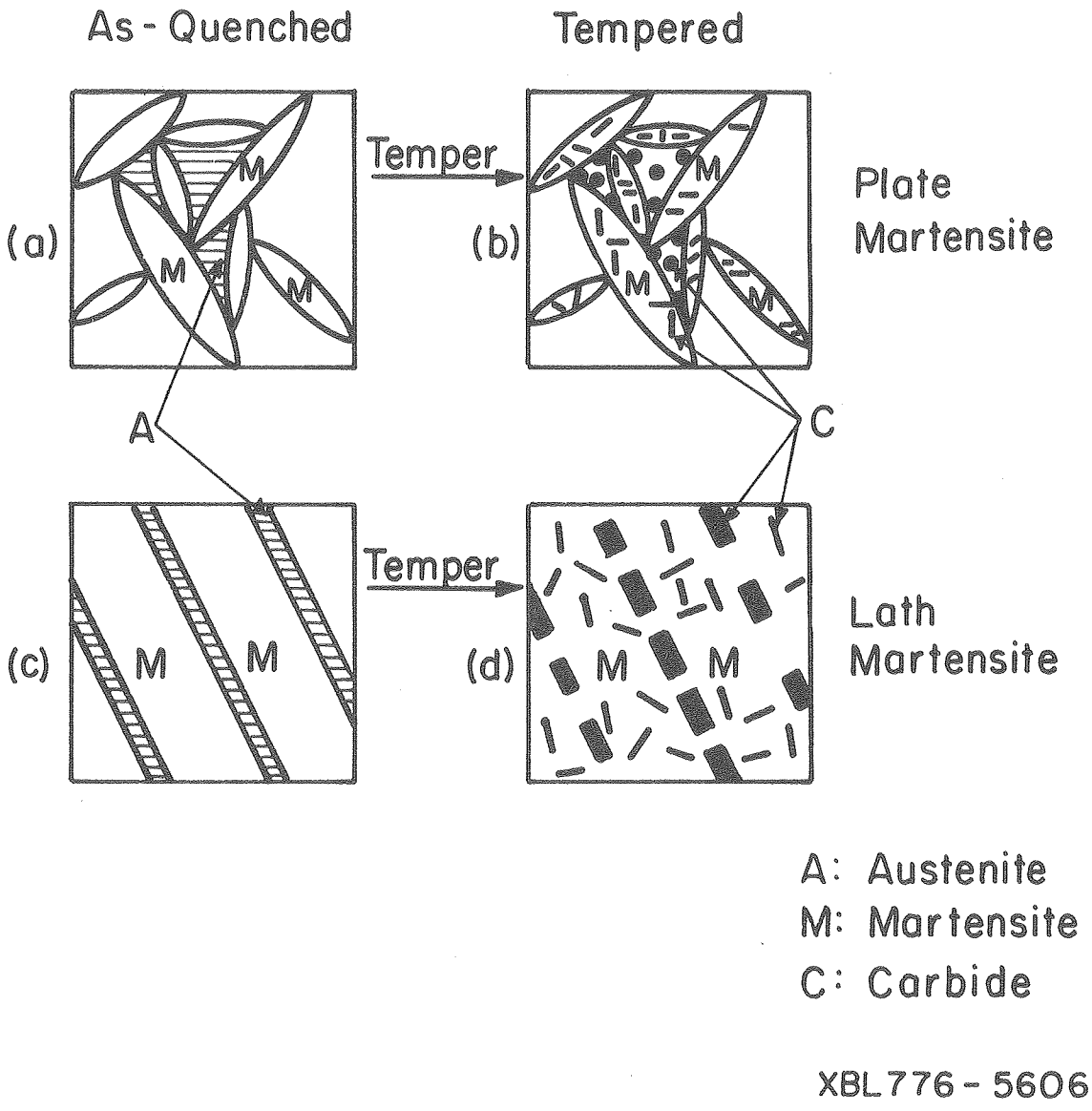

As - Quenched
Tempered
Temper
(a)
(b)
M
Plate
Martensite
A
C
Temper
(c)
(d)
Lath
Martensite
A: Austenite
M: Martensite
C: Carbide
XBL776 - 5606

Fig. 5

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