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FATIGUE PROPERTIES OF QUATOUGH STEELS
Fe/4Cr/XMn/0.25C

C.Y. Kung

October 1981

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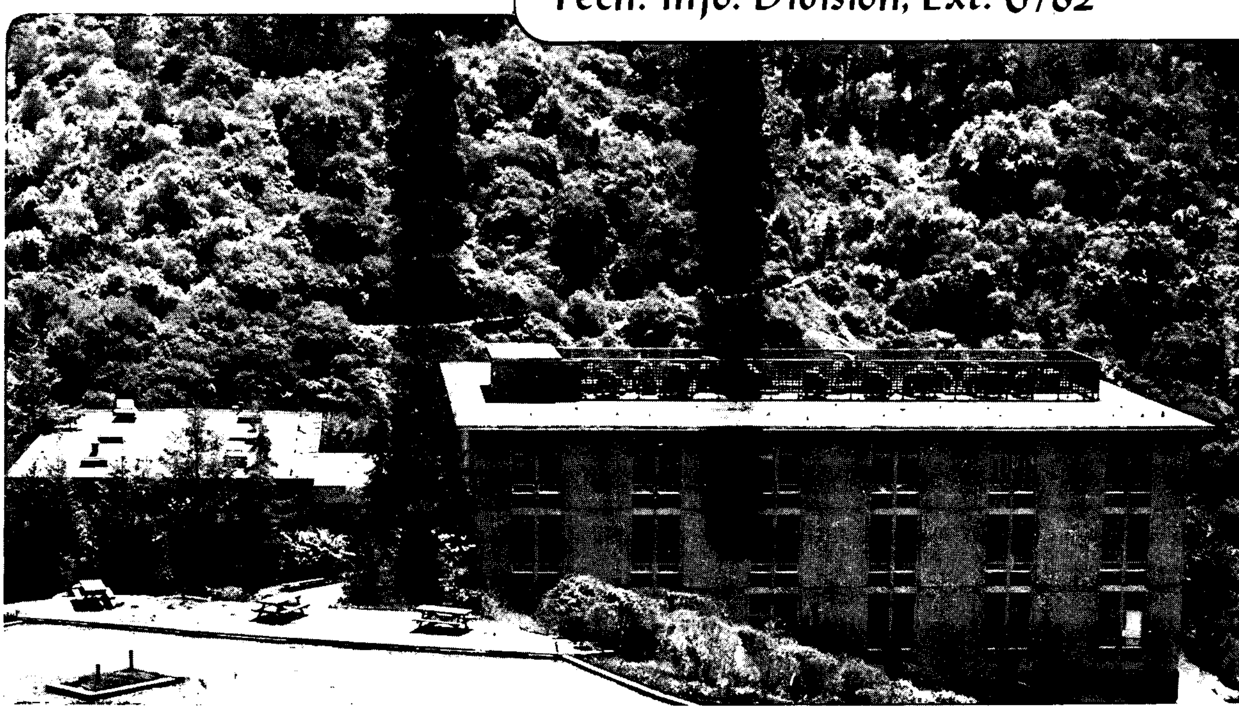

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Fatigue Properties of Quatough Steels Fe/4Cr/XMn/0.25C

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Introduction

Over the last decade at Berkeley, the low alloy medium carbon martensite Fe/Cr/Mn/C system (called quatough steels) has been developed with good combinations of strength and toughness[1,2]. Although both microstructures and monotonic mechanical properties have been well-studied[1,2], information is still lacking about fatigue properties of these steels. With the increasing potential application for quatough steels in a 'dynamic' structure (where the resistance to fatigue plays an important role), such as machinery in the mining industry, the fatigue properties of these newly developed steels must be carefully analyzed to avoid the possibility that an increase in toughness is obtained at the expense of fatigue properties. The scope of this study is to report the fatigue crack propagation results of the quatough steels in comparison to those of widely-used commercial high strength steels 4340 and 300M at the commonly used heat treatment conditions (i.e. as quenched + 200°C tempering).

It is well known that for high strength steels with yield strength over 1300MPa, the fatigue strength is dependent on ductility and toughness[3]. Ritchie and Knott[4] have also pointed out that static fracture modes are important for the fatigue fracture mechanism of the materials with low fracture toughness. It is reasonable to expect that static fracture parameters such as K_{1C} and E_{CV} values

(Charpy impact energy) would affect crack growth rate. Previous workers have derived theoretical and semi-empirical equations showing that the crack propagation rates are inversely proportional to K_{1C}[5,6] or (K_{1C}^2)[7-9] Among those, Miller[5] and Krafft[9] have tried to verify their equations from fatigue propagation results. However, the materials chosen for such verification exhibited a very large variation in strength. In certain tempering conditions, embrittlement induced by tempering can strongly affect the fatigue fracture modes. Therefore, these K_{IC} dependent formulae have not been properly verified by experiments. To isolate the effect of K_{IC} in the present study, a careful choice of material has been made. The Fe-4Cr-0.25C-XMn steels were selected and two different heat treatments were employed (as quenched, and as quenched + 200°C tempering) to obtain variations in toughness within the range of 65 - 195 MPa $m^{1/2}$ while maintaining at the same time a constant yield strength and similar microstructure. It would be interesting to establish any possible correlations between monotonic fracture resistance and cyclic fatigue fracture resistance. Such knowledge would lead to a better understanding and utilization of quatough steels.

Experimental Procedures

The quatough steels in this study nominally contained 4wt% Cr, 0.25wt% C, and 0 to 2wt% Mn. Materials were austenitized for one hour at 1100°C and then oil quenched. Subsequent tempering at 200°C for 1 hour was performed on the 0.5% and 2% Mn steels to obtain a higher toughness. Detailed composition and mechanical properties are listed in Table 1. The yield strength and hardness were found to be about the same for all steels whereas the K_{IC} and Charpy Impact Energy (E_{CV}) values varied significantly. The prior austenite grain size was about 110µm, and the microstructure consisted mainly of dislocated lath martensite with retained austenite. The amount of retained austenite was roughly proportional to the percentage of Mn in the material. Detailed descriptions of the microstructures of these steels have been thoroughly discussed elsewhere.[1,2]

Fatigue tests were performed on an MTS closed loop electroservohydraulic testing machine with a capacity of 20 kip (89kn). All tests were conducted at room temperature (~23℃) with approximately 45% relative humidity. The tests were carried out under stress-controlled conditions using a sinusoidal wave form (50 Hz) and R ratio (P_{min}/P_{max}) of 0.05.

The specimens used were 17.5mm thick compact tension steels with a slight modification in the notch depth as shown in Fig. 1. The specimen surfaces all underwent final polishing with a 1µm diamond paste in a direction parallel to the loading direction so as to minimize the scrathing effect on crack length measurement. Crack length measurements were determined by using a 30X travelling telemicroscope, calibrated to 10µm, with a readable accuracy of 2µm. The whole observation system gave the accuracy of about 0.1mm on absolute crack length which appeared on the specimen surface. The crack growth rates were studied in the range 0.45 - 0.85W (where W is the width of the specimen). All specimens were overloaded to final fracture. Scanning electron microscopy was employed to characterize the fracture morphology.

The grain boundary composition of intergranularly fractured areas was analyzed by using Auger spectroscopy to determine the amount of segregated impurities and alloying elements which were present. Fatigue fractured samples were carefully cleaned and examined using a primary electron beam of 20µm spot size.

The formulation of the stress intensity for the compact tension specimen used is as follows:[10]

$$K = \frac{(\Delta P) \cdot a^{1/2}}{W^{1/2} B} \left[29.6 - 185.5 \left(\frac{a}{W}\right) + 655.7 \left(\frac{a}{W}\right)^2 - 1017.0 \left(\frac{a}{W}\right)^3 + 638.9 \left(\frac{a}{W}\right)^4 \right]$$

where

K = stress intensity amplitude

ΔP = applied load amplitude = $P_{max} - P_{min}$.

B = specimen thickness

W = specimen width

a = crack length

Results and Discussion

Fatigue crack propagation rates. Fig. 2 shows the fatigue crack propagation rate results for the quatough steels, both A.Q. (as-quenched) and A.Q. + 200°C tempering. In the medium growth rate region (da/dN > 5 x 10^{-6}mm/cycle), the growth rate curves follow the Paris Law[11,12] with slope (m) around 2.8. The results of six different specimens all lay within a very narrow band with no obvious ordering with respect to toughness. Nowhere in this range of growth rates did crack growth rates vary by more than a factor of two. A recent ASTM-24 committed report[13] documented that crack growth rates for the same material, laboratory-tested under identical conditions, will vary typically by a factor of two. It can be concluded therefore that fatigue crack propagation responses at medium ΔK for the materials in this study were essentially the same. Accordingly, the K_{1C} dependent growth rate formulae (Kraft and Miller) are invalidated for the present case.

The experimental data show a scatter of about a factor of four or more at low ΔK region (da/dN > 10^{-6}mm/cycle). Two of the test results showed a retardation of growth at low ΔK. Such retardation phenomena which had been ascribed by former investigators to the crack closure effect[14] and the asymmetric development[15] of fatigue crack front at low ΔK, is the main cause for the extensive scattering observed. Unfortunately, this scatter in growth rate renders accurate observation of the present system difficult at a very low growth rate region (da/dN ~ 10^{-7} mm/cycle), Therefore, we were unable to obtain an accurate threshold stress intensity value, ΔK_0. However, by extrapolating the results of propagation rate data to the threshold region (as shown in Fig. 3), the ΔK_0 for quatough steels could be approximately estimated to lie in the range 2.9MPa $m^{1/2}$ to 4.5MPa $m^{1/2}$.

In the medium growth rate region, the present fatigue results show similar growth rates as compared to the earlier data of Fe-4Cr-0.35C ternary steels tested by Carlson and Ritchie,[16] as shown in Fig. 3. Fig. 3 also shows the growth rates of the quatough steels in addition to the other quenched and tempered steels 4340 and 300M. As the figure shows, the quatough steels displayed fatigue crack propagation rates that were appreciably less than those reported for the other steels at medium values of ΔK.[17,18]

Fatigue fracture morphology. Typical fracture surfaces for quatough steels at three different ΔKs, namely, 7MPa $m^{1/2}$, 15MPa $m^{1/2}$ and 25MPa $m^{1/2}$ are shown in Figs. 4a,b, 5a,b, and 6a,b, respectively. All "a" pictures show 2% Mn alloying steels at higher magnification, and "b" pictures show that the Fe/4Cr/0.29C base steel at low magnification. All fracture surfaces in this medium ΔK region (7MPa $m^{1/2}$ to 25MPa $m^{1/2}$) were found to be the same, independent of the composition of the steel. These fracture surfaces consisted mainly of transgranular cleavages with striations, in addition to intergranular cleavage facets. From the low magnification micrographs, there is an obvious difference in the amount of prior austenite intergranular cleavage surface present with respect to the variation in ΔK.

Experiments performed on 4340 steels by Miller[5,19] reported that materials which yielded slopes of between 2 and 3 were almost entirely propagated by striation growth. In our present experiments, the crack growth created a slope of

2.8. Striations mixed with intergranular facets were found. However, the interspacing between striations were found to be much larger than the crack length advanced per cycle. The fact that they were independent of ΔK as well as growth rate indicates that these were not likely to be classical fatigue striations.

The striation interval (1μm) was observed to be about the same order of magnitude as the martensite thickness; thus, fracture could have been transgranular with respect to prior austenite, but intergranular (or transgranular) with respect to individual laths.

Figs. 7a,b show the overall picture of the fracture surface with emphasis on the distribution of intergranular fracture surface. The area of intergranular facets increased as ΔK increased. However, for ΔK value of above 20MPa $m^{1/2}$, a reduction in intergranular facet was observed. The highest concentration was noted in the central part of the specimen.

An environmental condition such as water vapor has been proposed previously to have caused such intergranular facets.[20] However, if the environment had been a dominating factor, one would expect to find a larger percentage of intergranular embrittlement at the lower crack growth region where the reaction time of the material with environment is lengthened by the slower growth rate. Moreover, increased intergranular faceting would occur. No such occurrences were observed in the present work.

It is also suggested that intergranular fracture could have been induced by the segregation of Cr or certain residual impurity elements (such as S, P, Sb, Sn, etc.) along grain boundaries at the central part of the specimen where the cooling rate was lower.[21] To verify such a hypothesis, Auger electron microscopy was performed at different areas of the specimen. Spectrum 1 in Fig. 8 shows the chemical concentration in an intergranular fracture surface. Spectrum 2 and 3 were taken from regions with regular fracture surfaces. No measurable partitioning was observed of either Cr or any impurity elements. It is likely, therefore, that segregation did not play a major role in intergranular fracture. The high sulphur concentration on all fracture is believed to be the result of contamination produced when the Auger specimens were cut from the fatigued sample.

The presence of intergranular facets cannot be attributed to the modes of static or monotonic fracture, since such intergranular facets were not seen in K_{1C} test (monotonic fracture test) samples at any value below K_{1C} in the present experiment. This also suggests that little correlation can be established between the monotonic fracture resistance and fatigue resistance parameters for the K range studied. This agrees with the present experimental findings in which the propagation rate was independent on K_{1C}.

Secondary cracking took place during fatigue crack propagation in the quatough steels at medium and high stress intensity levels. It was relatively high for 2wt% Mn steels both A.Q. and A.Q. + 200°C tempering conditions, as shown in Fig. 6a. These secondary cracks could be caused by crack branching in regions with a continuous film of retained austenite. This branching is believed to increase the monotonic fracture resistance (K_{1C} and E_{CV} values) and was expected to retard the fatigue crack growth rates by partitioning the stress intensity. Nevertheless, in the present results, the variations of secondary crack density did not lead to a detectable difference in fatigue crack growth rates.

Tensile Tear Fracture Morphology. Figs. 9a,b,c show the fracture morphology of overload fracture. They are strange as compared to K_{1C} fracture surfaces. At the onset of tensile fracture (a/w ~ 0.85), the monotonic morphology was comprised mainly of ductile dimples (A) and a vein-like structures (B). Occasionally, cleavage fractures (C) and rib-like patterns (D) were also seen. For the rib-like pattern in Fig. 9a, inclusions of tiny particles of about 0.2μm were found between the striations of the ribs. This inclusion-induced rib-like pattern could be formed as a result of coalescence of micro voids in the vicinity of inclusions. At crack length around 0.9W, the rib-like pattern becomes the major feature. No inclusion has been associated with it. This kind of pattern has been reported in 2%Ni quaternary plus alloys in a 300℃ tempered CVN (Charpy V-notch) sample.[22] The majority of the striation orientation was along directions perpendicular to the crack extension direction. The underlying reason for such arrangement is still not fully understood and needs further investigation.

Conclusions

From studies of fatigue crack propagation in the quatough steels, tested at a constant strength and with a similar microstructure at room temperature under moist air, the following conclusions can be made:

(1) Variations in fracture toughness (K_{1C}) from 65MPa $m^{1/2}$ to 198MPa $m^{1/2}$, induced by Mn alloying element and 200℃ tempering, have no effect on the "mid-range" of crack growth rates exceeding 5×10^{-6} mm/cycle. This finding is inconsistent with the empirical formulae which describe an inverse proportional relationship between da/dN and K_{1C}.

(2) Quatough steels have better fatigue resistance in the medium ΔK range as compared to the commercial steels AISI 4340 and 300M at the same heat treatment condition (A.Q. and 200℃ tempering).

(3) Retained austenite, which has been shown to have a beneficial effect on the monotonic fracture toughness of the quatough steels, is observed to have little or no effect on the crack growth rate of these steels, in the mid-range growth rates, consistent with the lack of microstructural influence on crack propagation generally observed for steels in this growth rate region.

Acknowledgements

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Figure Captions

Fig. 1. Modified compact tension specimen for fatigue test.

Fig. 2. Fatigue crack propagation results in Fe-4Cr-0.3C-XMn steels.

Fig. 3. Comparisons of fatigue crack propagation rates vs. stress intensity range of quatough steels with rates reported for other high strength steels.

Fig. 4. Typical fracture surface for quatough steels at $K \sim 7MPa\ m^{1/2}$ (a) for 2%Mn alloying steel; (b) for ternary base steel.

Fig. 5. Typical fracture surface for quatough steels at $K \sim 13MPa\ m^{1/2}$ (a) for 2%Mn alloying steel; (b) for ternary base steel.

Fig. 6. Typical fracture surface for quatough steels at $K \sim 23MPa\ m^{1/2}$ (a) for 2%Mn alloying steel; (b) for ternary base steel.

Fig. 7. Showing the distribution of intergranular fracture surface with respect to K.

Fig. 8. Auger spectrum from intergranular fracture surface and transgranular cleavage surface of fatigued quatough steels.

Fig. 9. Representative SEM fractographs of quatough steels specimen fractured by overload at room temperature.

Table 1. Mechanical Properties of Quatough Steels

Alloy	Tempering Temperature	YS Ksi (MPa)	UTS Ksi (MPa)	K_{1C} Ksi in$^{1/2}$ (MPa m$^{1/2}$)	E_{CV} ft-lbs (NM)	V.R.A. (%)
Fe/4Cr/0.29C	A.Q.	195 (1371)	240 (1687)	71.0 (78)	18.0 (24.5)	
Fe/4Cr/0.24C/0.54Mn	A.Q.	192 (1350)	233 (1638)	92.5 (102)	14.0 (19)	0.7
	A.Q.+200°C	175 (1230)	214 (1504)	150.0* (165)	35.0 (47.6)	0.7
Fe/4.1Cr/0.24C/1.0Mn	A.Q.	198 (1392)	240 (1687)	95.5 (94)	11.5 (15.6)	1.1
Fe/4Cr/0.25C/2.0Mn	A.Q.	207 (1455)	265 (1863)	59.0 (65)	3.4 (4.6)	1.7
	A.Q.+200°C	195 (1371)	235 (1652)	180.0* (198)	40.0 (54.4)	1.7

Specimens were austenitized at 1100°C for 1 hour. A.Q. = as quenched.

YS = 0.2% off yield strength

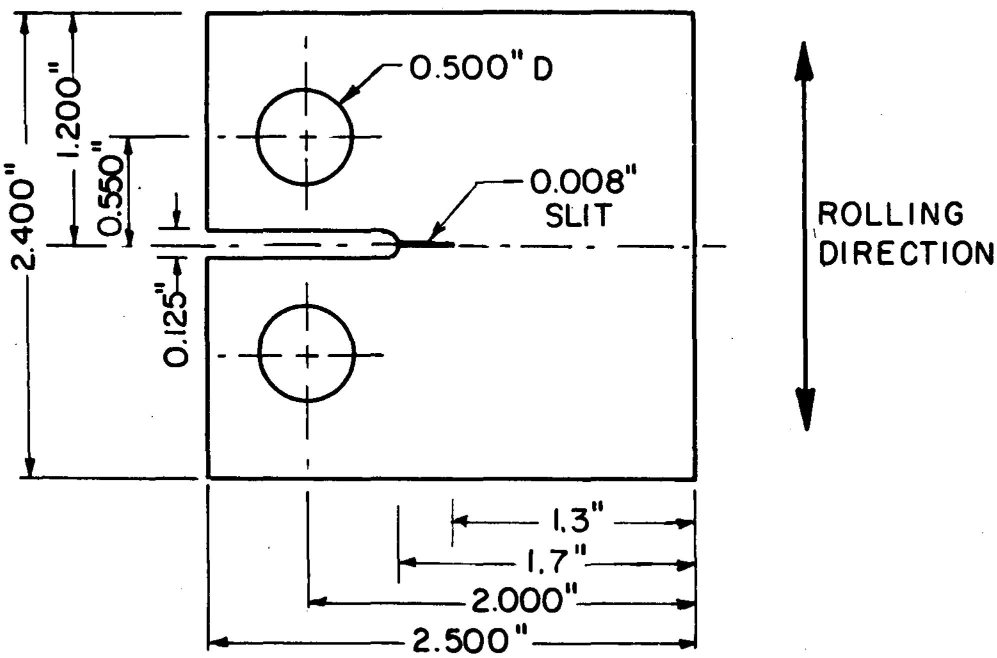
UTS = ultimate tensile strength

K_{1C} = fracture toughness

E_{CV} = Charpy Impact energy

V.R.A. = volume fraction of retained austenite

*Violated plane strain conditions, calculated K_Q instead of K_{1C}.



FRACTURE TOUGHNESS SPECIMEN

XBL 754-6176A

Fig. 1

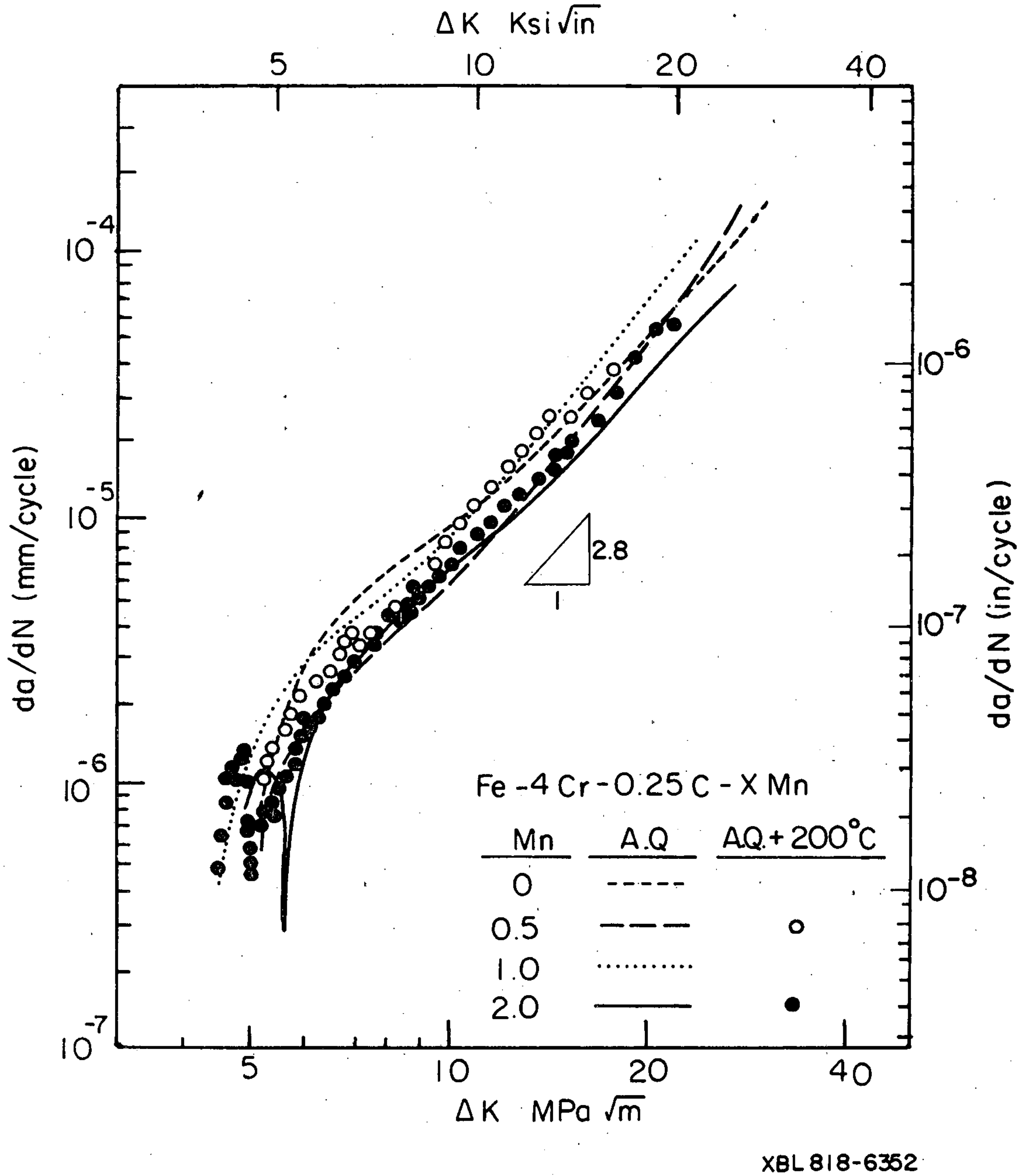

ΔK Ksi√in
5
10
20
40
10^{-4}
10^{-5}
10^{-6}
10^{-7}
da/dN (mm/cycle)
10^{-6}
10^{-7}
10^{-8}
da/dN (in/cycle)
2.8
1
Fe -4 Cr - 0.25 C - X Mn
Mn
A.Q
AQ + 200°C
0
0.5
1.0
2.0
5
10
20
40
ΔK MPa √m
XBL 818-6352

Fig. 2

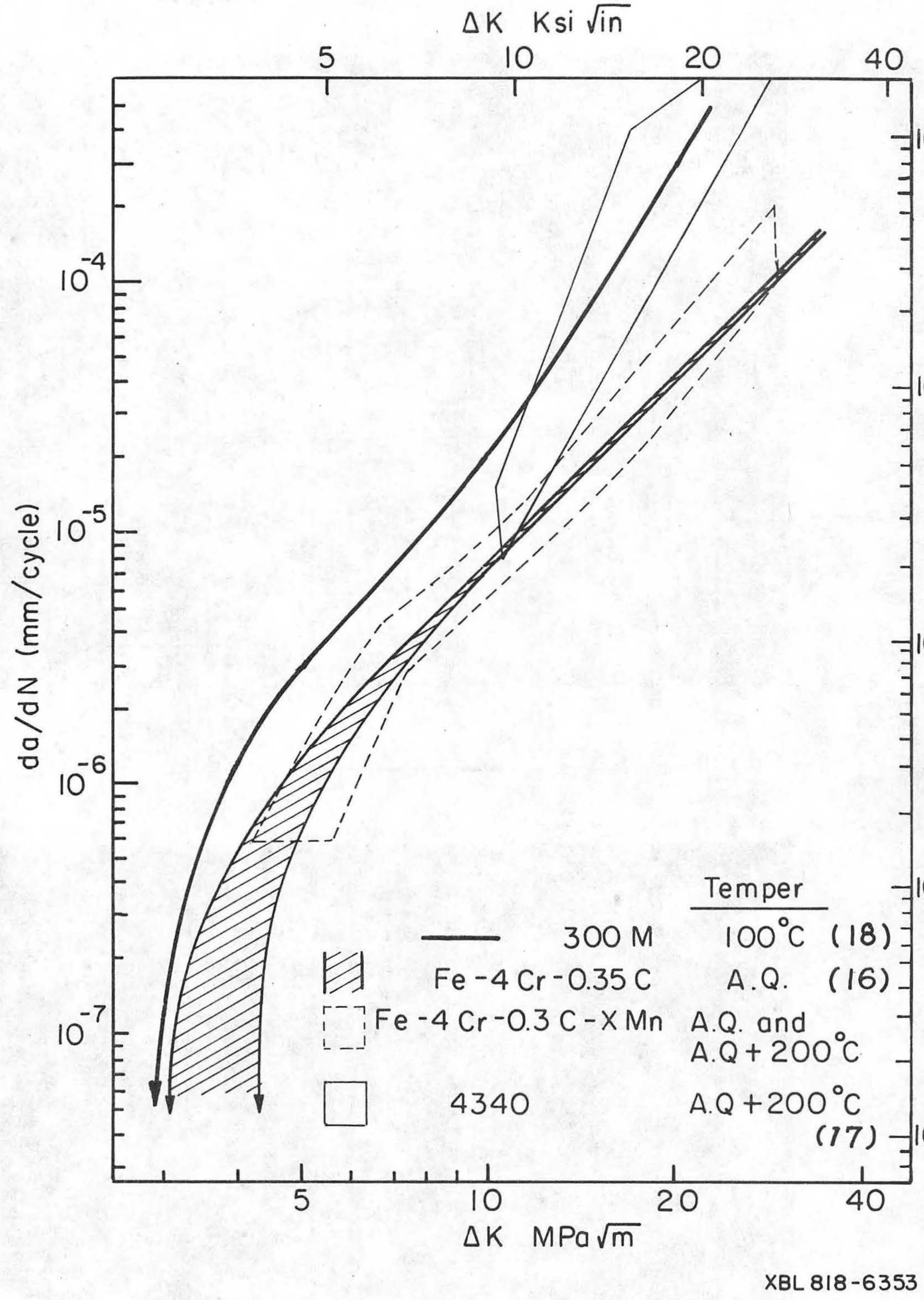

ΔK Ksi √in
5
10
20
40
da/dN (mm/cycle)
10^{-4}
10^{-5}
10^{-6}
10^{-7}
da/dN (in/cycle)
10^{-5}
10^{-6}
10^{-7}
10^{-8}
10^{-9}
Temper
300 M
100°C (18)
Fe -4 Cr -0.35 C
A.Q. (16)
Fe -4 Cr -0.3 C - X Mn
A.Q. and
A.Q + 200°C
4340
A.Q + 200°C
(17)
5
10
20
40
ΔK MPa √m
XBL 818-6353

Fig. 3

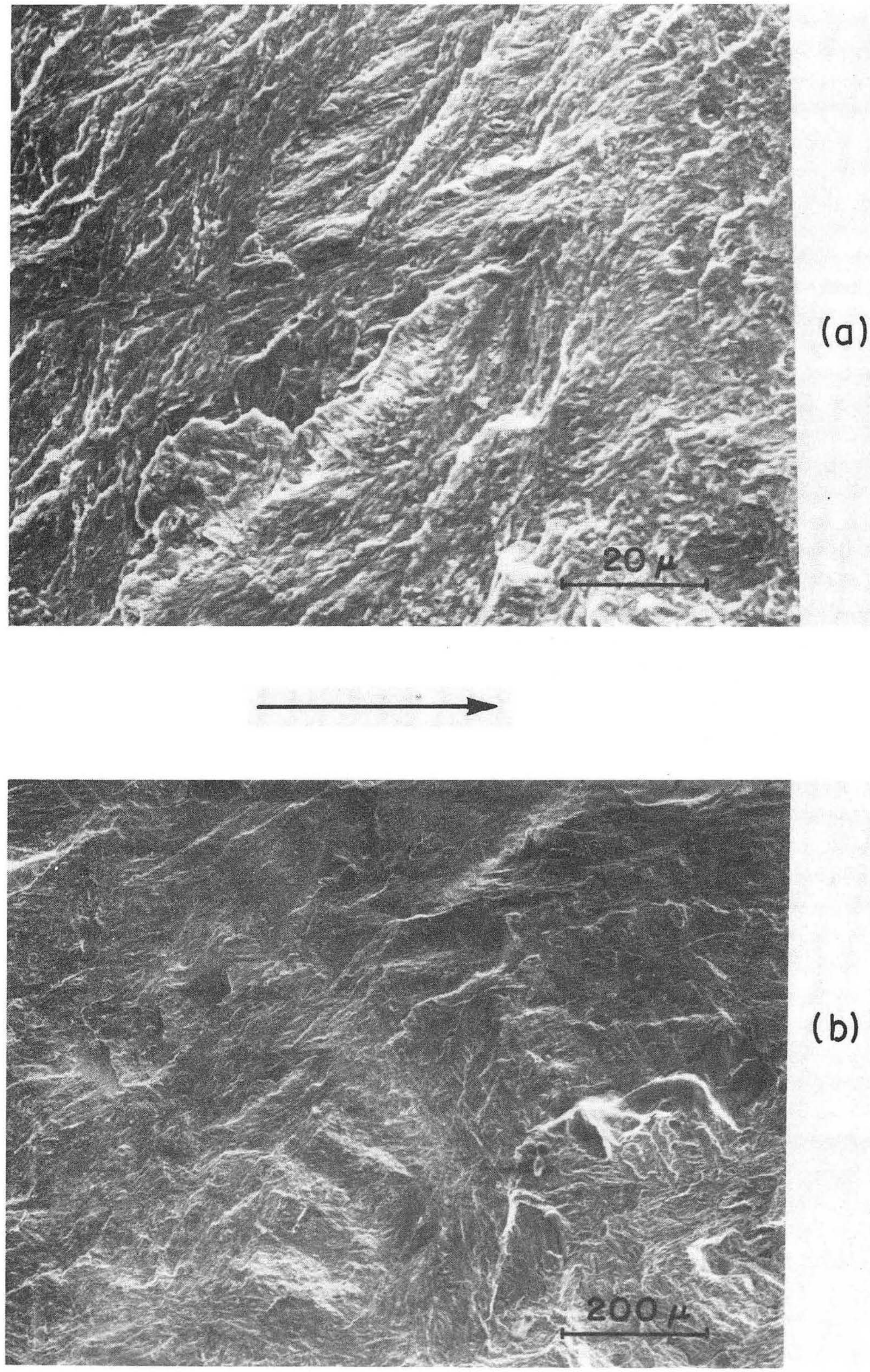


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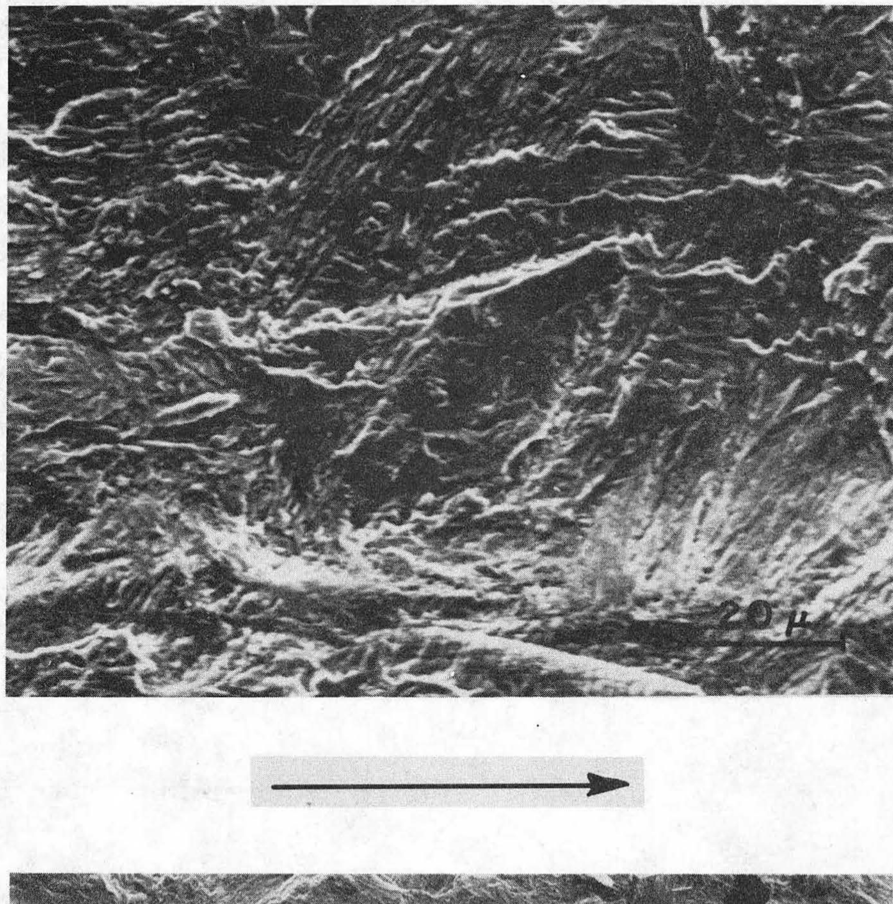
Fig. 4



(a)

(b)

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Fig. 5

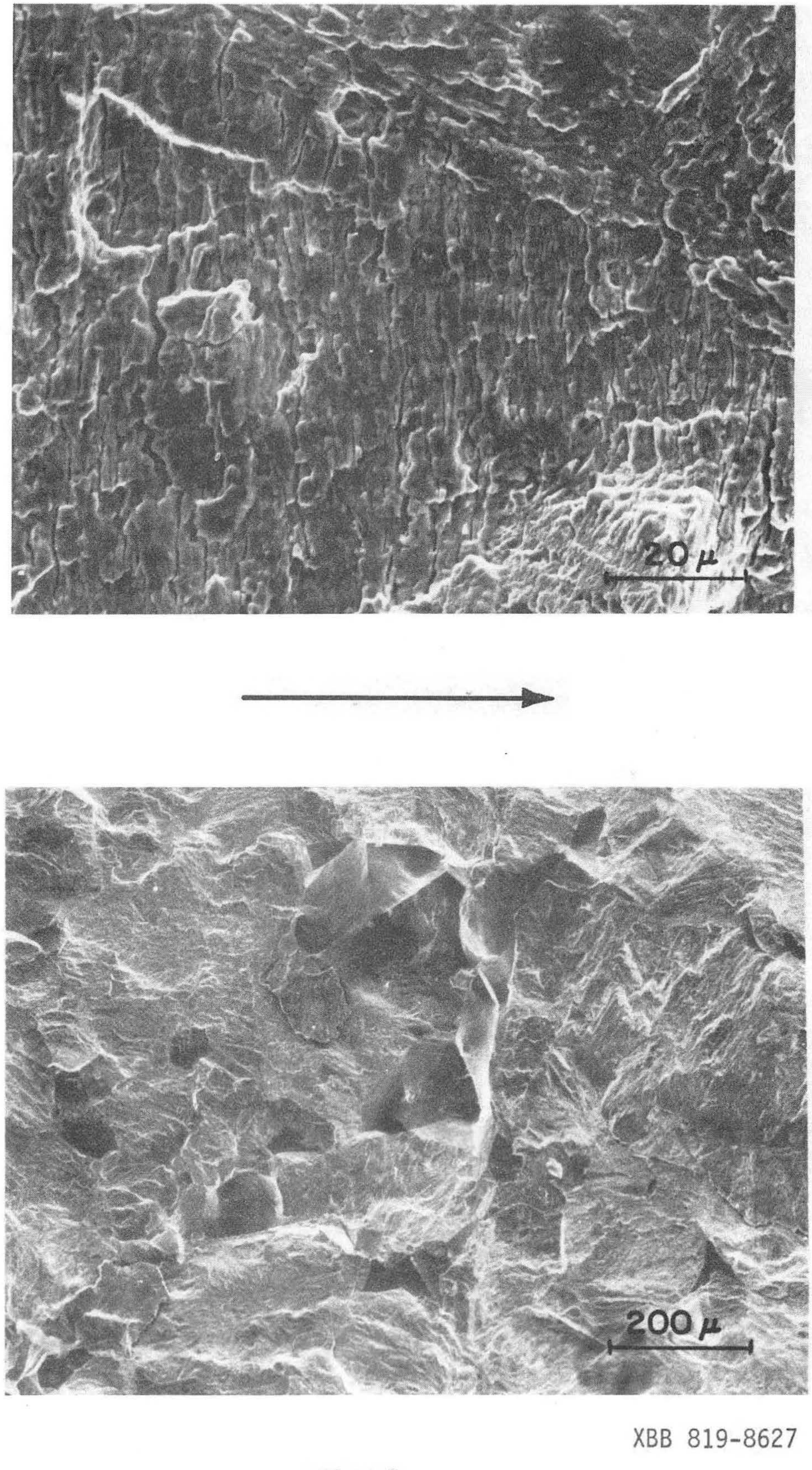

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Fig. 6

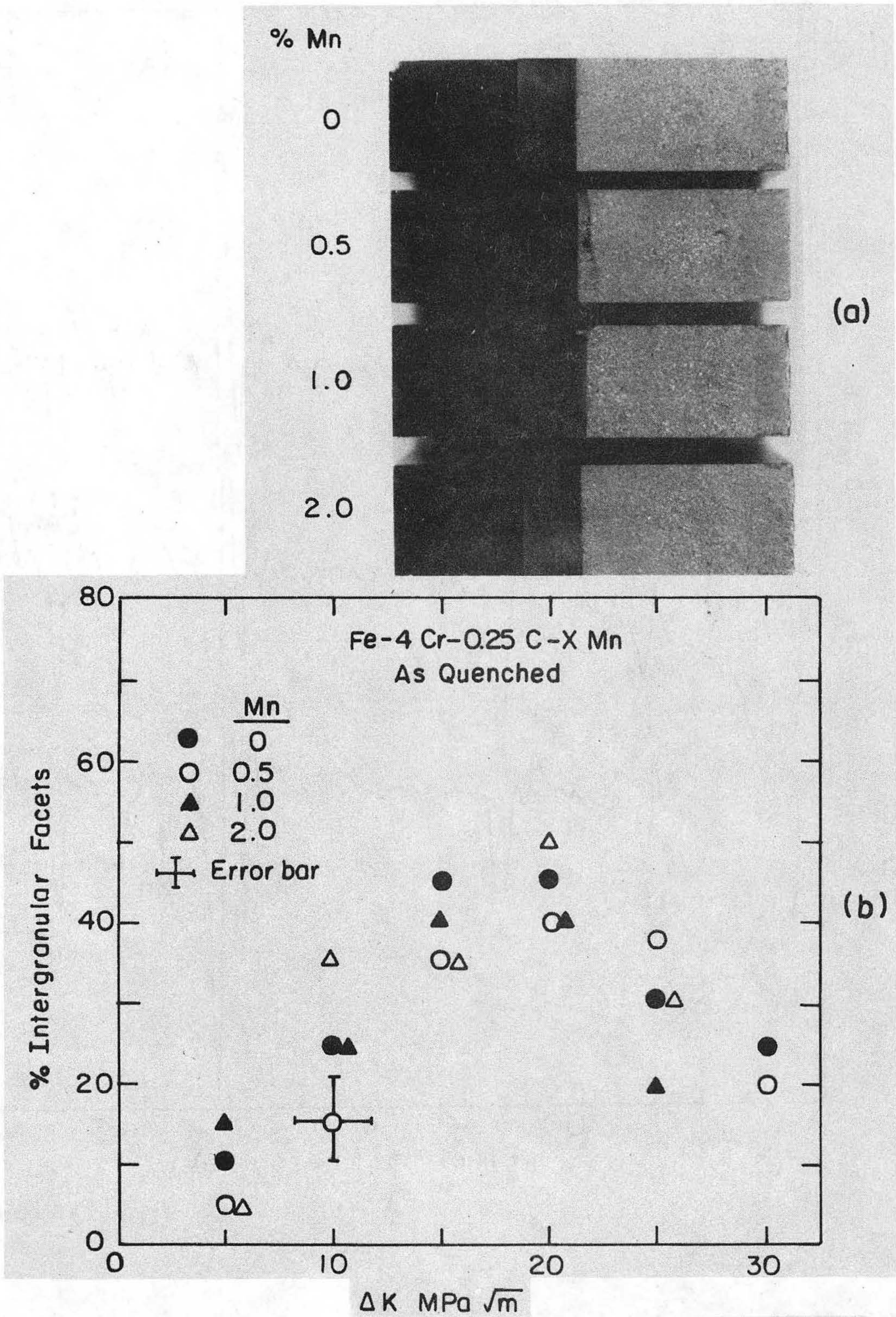

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Fig. 7

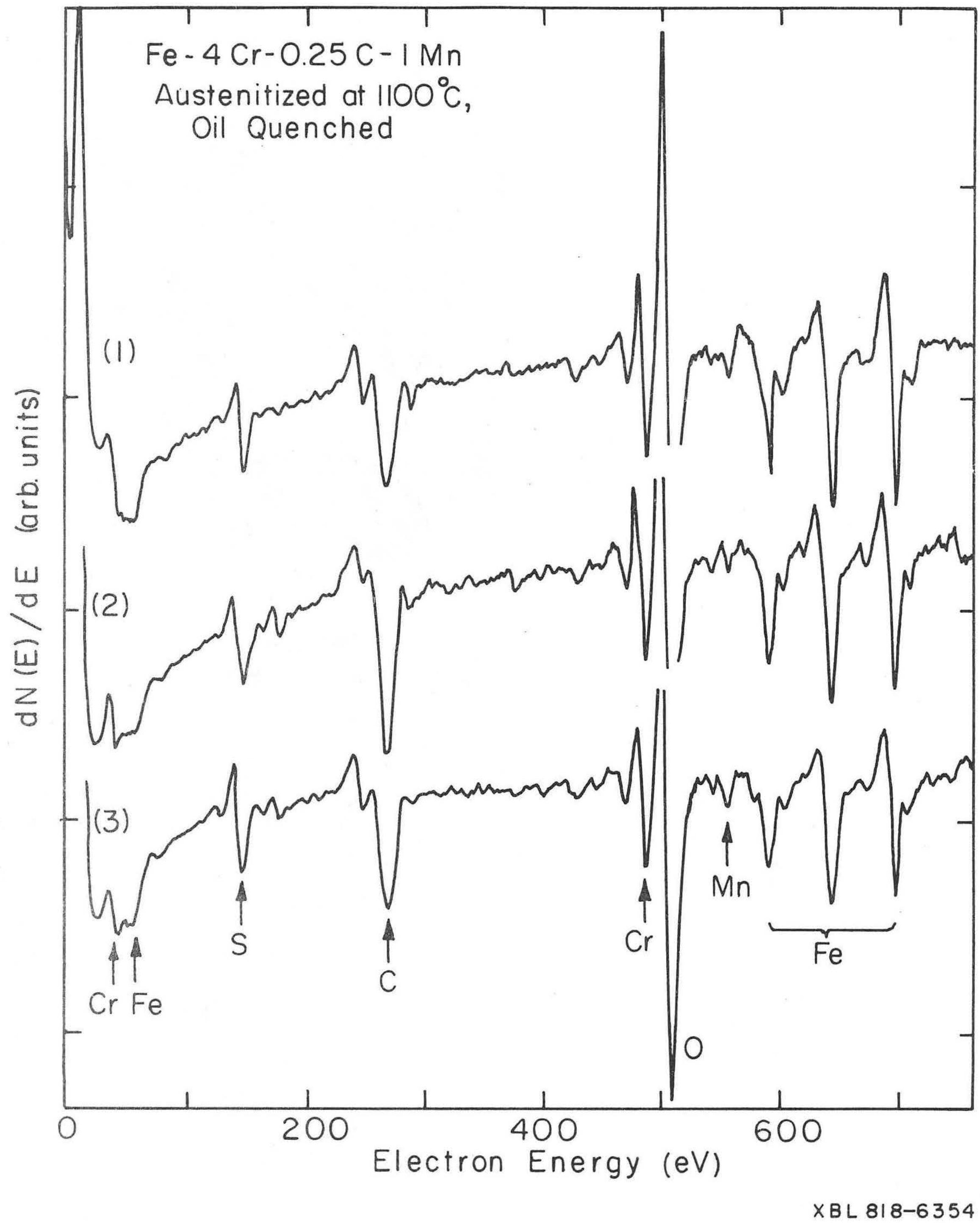
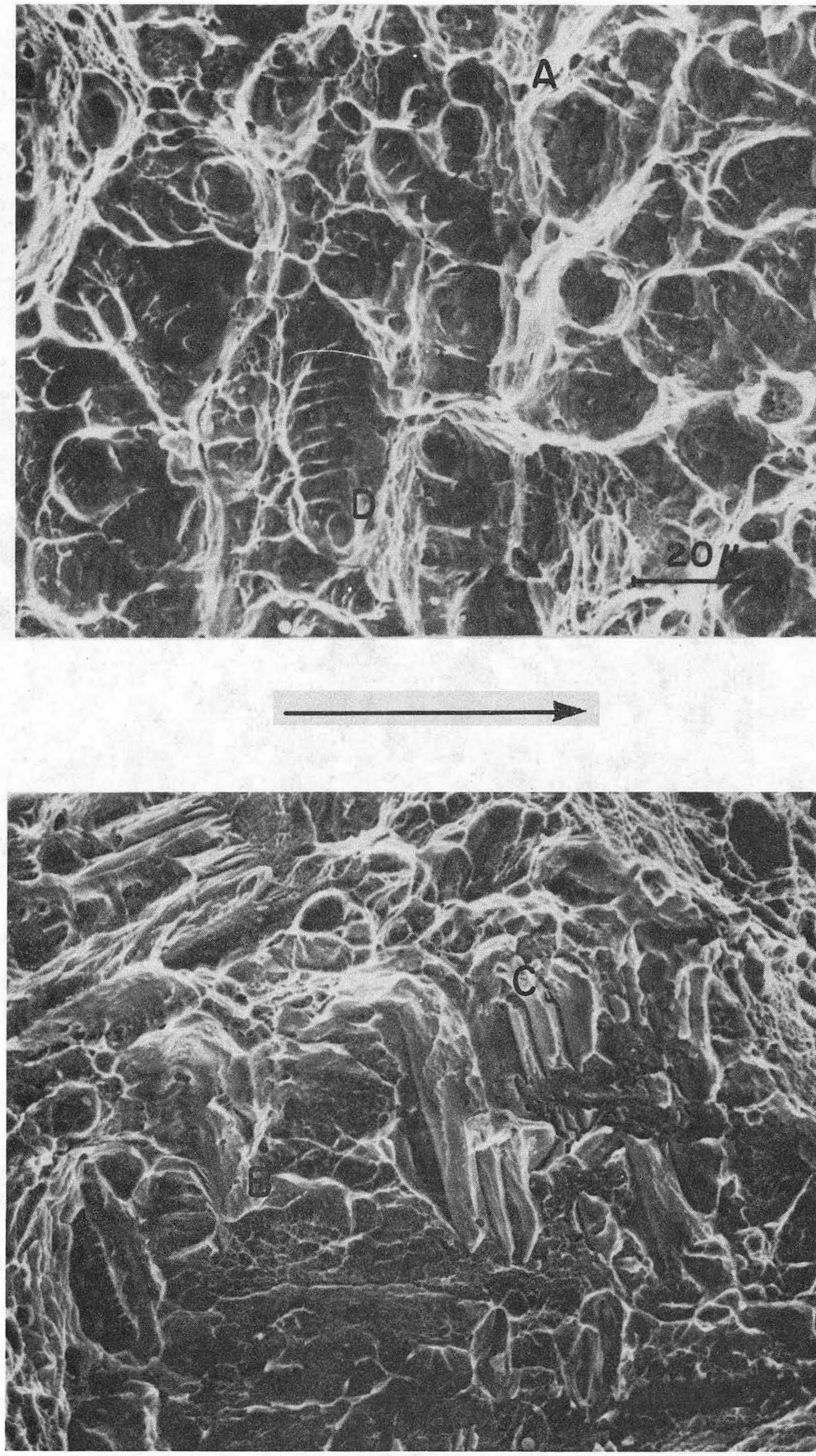

Fe-4 Cr-0.25 C-1 Mn
Austenitized at 1100°C,
Oil Quenched
(1)
(2)
(3)
dN(E)/dE (arb. units)
S
C
Cr
Cr Fe
Mn
Fe
O
0
200
400
600
Electron Energy (eV)
XBL 818-6354

Fig. 8



(a)

(b)

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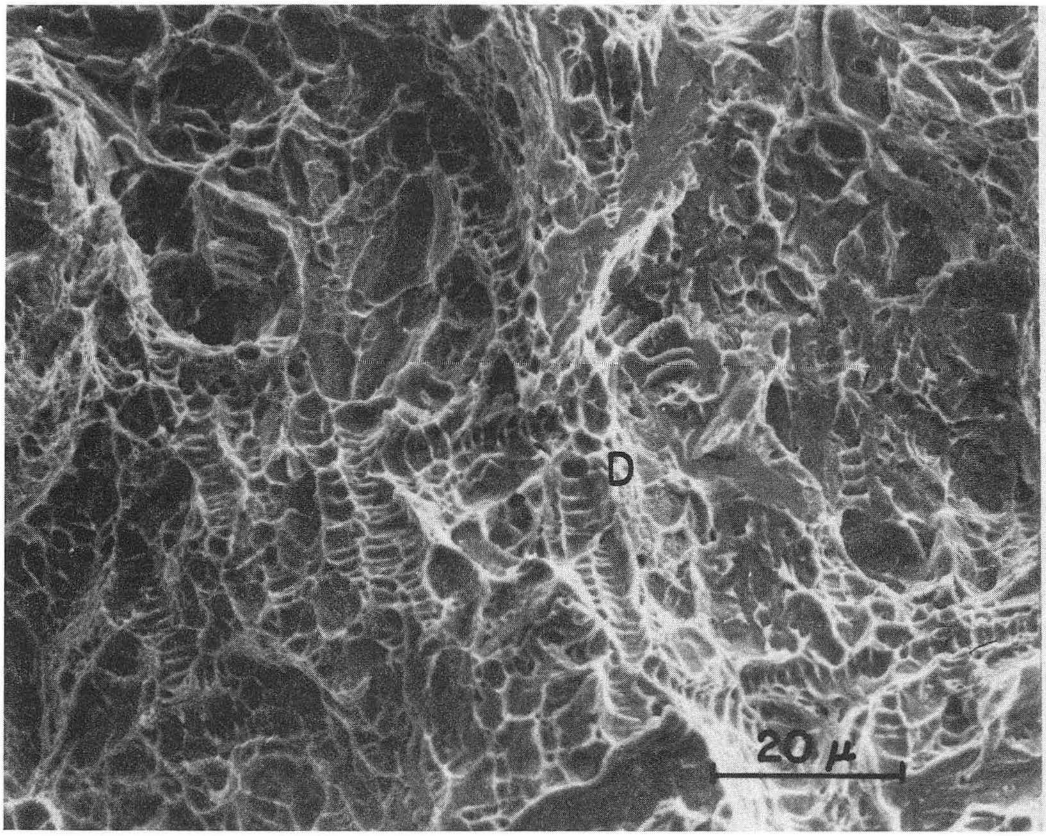
Fig. 9



(c)

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Fig. 9

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