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FATIGUE CRACK GROWTH RATES OF
A SECONDARY HARDENING STEEL

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(M. S. thesis)

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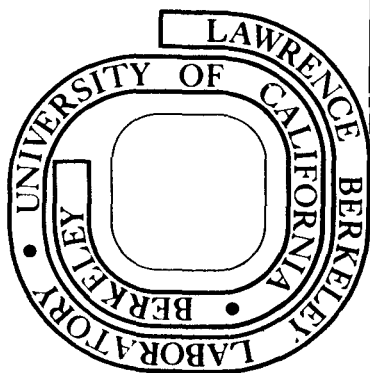
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I. INTRODUCTION

Fatigue crack propagation behavior in high strength steels is very important in determining the usefulness of such steels when they are subjected to cyclic loading. The crack initiation period, which is given considerable importance in conventional fatigue studies, is often absent in the high strength materials because of pre-existing flaws, such as inclusions. Linear elastic fracture mechanics has been applied to crack propagation because of its successful use in predicting fracture conditions. Paris and his colleagues^{1,2} have proposed an empirical equation relating the stress intensity, K , to crack growth rate. They showed that the growth rate was related to the stress intensity range in the vicinity of the crack tip. Their relation is

$$\frac{da}{dN} = C_0 (\Delta K)^m$$

where da/dN is the change in crack length per cycle of loading, ΔK is the range of stress intensity being applied during cyclic loading, and C_0 and m are experimental constants. This relationship has been used extensively to characterize the fatigue crack growth behavior of a variety of materials.³⁻⁸ The relation has also been used extensively by investigators to show the dependence of crack growth rates in certain materials on different variables. In particular, fatigue crack growth behavior in quenched and tempered steels has been shown to depend on several factors. Steel cleanliness, i.e., impurity levels

can markedly effect the crack growth rates.⁹ Specimen thickness, which dictates whether the steel is under plane strain or not has been shown to influence crack growth rates.¹⁰ The influence of tempering treatment has also been investigated in some steels, leading to attempts to correlate fatigue crack growth behavior with yield strength as well as other mechanical properties.^{11,12} Finally, there has been extensive investigation of the influence of environment on fatigue crack growth properties, particularly in the quenched and tempered steels.¹³ Several investigators have shown that moisture levels in air have an influence on the fatigue crack growth properties.^{9,14} It has been shown, too, that other variable influences, such as impurity levels, are exaggerated in a moist air environment.

Although these many studies have identified the variables that influence fatigue crack growth rates, there have been few attempts to relate the actual microstructure present to the fatigue crack growth behavior. The microstructural changes accomplished by tempering are often small. However, in a secondary hardening steel there are drastic changes that occur at the higher tempering temperatures.¹⁵ Austenitizing temperature can also affect the microstructure drastically by influencing cooling rates and decomposition products formed during initial quenching.¹⁶

Previous fracture toughness work has shown that large variations in toughness at equivalent yield strength levels occur in a quenched and tempered 5Mo-.3C secondary hardening steel.¹⁷ The microstructural changes resulting from different tempering treatments were characterized

by transmission electron microscopy and used to explain the observed variation in fracture toughness. A similar study of a series of Mo-C secondary hardenings steels was also successful in explaining the role of austenitizing temperature on fracture toughness.¹⁸ The purpose of the present study was to investigate the fatigue crack growth properties in the 5Mo-.3C steel. The other mechanical properties for different austenitizing treatments as well as for the different tempering treatments were determined in the previous study. The major objective in the present study was to determine how the microstructural features that controlled the previously measured fracture toughness and tensile properties affected the fatigue crack growth behavior. The study's purpose was to investigate the fatigue crack propagation rates in an air environment to ascertain the material's behavior in a normal environments. Studies in argon would have helped explain environment - microstructure interaction, but the extension of crack growth studied in this area was not possible because of time limitations.

II. EXPERIMENTAL PROCEDURE

A. Material Preparation

The steel used for this investigation was melted in a vacuum induction furnace. The majority of the material of approximately 5Mo-.3C composition came from three 20 lb. ingots, whose chemical analyses are given in Table I. Each ingot was homogenized for 24 hours at 1100°C and then hot-rolled at 1100°C into bars of width 2 3/4 in and thickness 5/8 in. Material of two other compositions, given in Table I, was also used. This material was first hot-forged at 1100°C, hot rolled to thickness at 1100°C, and then homogenized at 1200°C in vacuum for 48 hours.

Specimens for mechanical testing were cut from the rolled bars, and then austenitized for one hour in a vertical tube furnace with an argon atmosphere. The majority of specimens were austenitized at 870°C, or 1200°C, with one specimen also austenitized at 980°C. Following austenitization, the specimens were quenched in agitated ice-brine, and then refrigerated in liquid nitrogen. The tempered specimens were held for one hour at the desired temperature in a salt bath. The tempering temperatures used were 225°C, 300°C, 500°C, and 600°C.

B. Mechanical Testing

1. Hardness Testing

Vickers microhardness values of heat treated specimens were determined using a Leitz Wetzlar microhardness testing unit. An applied load of 1000 g was used in measuring the hardness of the specimens from metallographic samples taken from fatigue crack specimens. Both indentation diagonals were measured and the average value of hardness was recorded. Five tests were made on each specimen to determine the average microhardness.

2. Tensile Tests

Tensile properties were determined using the 1 in. gage length, 1/4 in. diameter round specimen shown in Fig. 1. The specimens were rough machined oversize, heat treated, and ground to the final dimensions. Tests were conducted at room temperature using a 300,000 pound capacity MTS, at a loading rate of 0.04 in/min.

3. Fatigue Crack Growth Properties

Fatigue crack propagation properties were measured using a compact tension specimen shown in Fig. 2. The dimensions used comply with the ASTM standard for specimens used to determine plane strain fracture toughness.¹⁹ The thickness used was 9/16 in. The crack-line loaded specimens were machined to the dimensions displayed prior to heat treatment. After heat treatment, the sides were wet ground to remove approximately 0.01 in. and the 0.008 in. slot at the tip of the notch was cut.

The fatigue crack growth tests were carried out at room temperature (20-25°C) in air (45-55% relative humidity) using a 300,000 lb. capacity MTS machine. The specimens were tested at a frequency of 6Hz under sinusoidal tension-tension cycling. Prior to monitoring crack advancements, a fatigue pre-crack was initiated in accordance with ASTM specifications.¹⁹ Once the crack had been initiated, fatigue crack advancements were measured optically using a traveling microscope, as shown in Fig. 3. Increases in crack advancement of 0.001 in. could be measured on both surfaces of the specimen, and an average was used. The load was monitored continuously, and most tests were carried out using a minimum to maximum load ratio of $R = 0.1$ (where $R = P_{\min}/P_{\max}$). Although displacement was controlled using the short stroke function, the displacement was continuously changed to keep the loads constant. Therefore, the stress intensity range at the crack tip increased as the crack length increased. The formulation of the stress intensity for the compact tension specimen used is as follows:

$$K_I = \frac{P}{W^{1/2} B} \left[29.6 \left(\frac{a}{W} \right)^{1/2} - 185.5 \left(\frac{a}{W} \right)^{3/2} + 655.7 \left(\frac{a}{W} \right)^{5/2} - 1017.0 \left(\frac{a}{W} \right)^{7/2} + 638.9 \left(\frac{a}{W} \right)^{9/2} \right]$$

where a , the crack length, was measured from the center of the pinholes. Because the stress intensity range increased, many data points were collected for each specimen.

4. Fracture Toughness Testing

Fracture toughness values were determined using the fatigue crack growth specimens. When the a/W ratio reached about 0.7, a rising load test was conducted using the fatigue specimen. A 5000 kg capacity Instron testing machine was used to perform the tests at a loading rate of .1 cm/min. A crack-opening displacement (COD) gage similar to that described in the ASTM standards was used to monitor crack length during testing.¹⁹ The calibration curve for the COD gage is given in Fig. 4. Even though the tests did not meet ASTM specifications because of the a/W ratio, the fracture toughness values gave relative measures of fracture resistance.

C. Microscopy

1. Optical Metallography

Specimens for optical metallography were cut from the heat treated fatigue specimens. After mounting in Koldmount, the specimens were ground on silicon carbide papers to 600 grit, polished on a 1 μ diamond abrasive wheel, and given final polish in a system with 0.05 gamma alumina micropolish. Specimens were then etched with 2% to 5% nital solutions for 5 to 30 seconds.

2. Scanning Electron Microscopy

Both the fatigue and rising load areas of the fracture surfaces of each fatigue specimen were examined using a Jeolco JSM-U3 scanning electron microscope. The surface was protected with acetate tape during preparation for observation. Prior to insertion into the microscope the acetate tape was stripped off and the specimen was ultrasonically cleaned in acetone.

III. RESULTS AND DISCUSSION

A. Results and Discussion of Treatments Subsequent to Austenitization at 1200°C

1. Mechanical Properties

a. Hardness, tensile, and fracture toughness. The 5Mo-.3C steel used in this study is classified as a secondary hardening steel. This category of steels retains high room temperature strength after tempering at high temperatures following a quenching from an austenitizing treatment in which carbides have been dissolved. This behavior has been verified in this steel in previous studies.¹⁷

The hardness vs. tempering temperature curve for this steel is shown in Fig. 5. As expected, there is an increase in hardness in the neighborhood of 600°C, where a peak occurs. The tensile properties are listed in Table II, and visually displayed in Fig. 6. The yield strength behavior increases follows the hardness behavior.

The fracture toughness values with tempering temperature are listed in Table II, and shown in Fig. 7. The fracture toughness dropped significantly when tempering temperatures greater than 225°C were used. The lowest value, observed at the tempering temperature of 600°C, was similar to the value shown by Goolsby.¹⁷

b. Fatigue crack growth properties. The fatigue crack growth rates for varying ΔK values were determined for all tempering treatments. The data are listed in Tables III through VII. The Paris relation was used as a basis for the fatigue crack growth analysis and a least-squares log-log fit²⁰ was used to determine the parameters in the

empirical relation, as shown in Figs. 8 through 12.* Duplicate tests were conducted to determine the fatigue crack growth parameters for all tempering treatments, except the 600°C treatment.

In the following discussion an attempt is made to determine how fatigue crack growth is related to the microstructural features of the steel and to establish whether the same microstructural features affecting yield strength and fracture toughness affect fatigue crack growth properties in a similar way.

2. Microstructure and Crack Growth Behavior

a. As-Quenched steel. Figure 13 is an optical photomicrograph showing the martensitic structure present in the as-quenched material. The large austenite grain size, 300-400 μ diameter, is apparent. As shown by transmission electron microscopy by Goolsby, extensive autotempering was present in the martensite laths. Extensive autotempering occurred because the M_s temperature was high (about 425°C). The high M_s of the material was a consequence of the relatively low carbon content (0.3%) and the high austenitizing temperature. Goolsby has shown,¹⁷ using the dark field transmission electron microscopy techniques, that the only carbide present in the autotempered martensite was cementite. No epsilon carbide was detected. The martensite was predominantly the dislocated lath type, which is characteristic of steels of this carbon content.²¹ Internal twinning was observed only in isolated regions.

Figure 14 is a composite scanning electron micrograph of the fatigue fracture. The ΔK level in the area shown was 35 ksi-in.^{1/2}. The macroscopically measured crack growth rate at this ΔK was

* ΔK values entered as psi-in.^{1/2}

1.5×10^{-5} in./cycle. Features characteristic of quasi-cleavage were observed throughout the fatigue area. An example of this is shown in the fatigue area in Fig. 14 at point A. This quasi-cleavage feature may account for the scatter in some of the data. Surrounding the quasi-cleavage area, A in Fig. 14, is evidence of localized plastic instability (at B) on the fatigue fracture surface. The diffuse markings resembling striations (at B) also suggest that the fatigue crack was not confined to a single plane. At C in Fig. 14, indications of striations are also seen. Striations are revealed more clearly at A in Fig. 15, which is a scanning electron micrograph of another area where the ΔK was $40 \text{ ksi-in.}^{1/2}$. The measured average microscopic spacing of these striations (1.6×10^{-5} in.) was approximately equal to the macroscopically measured crack growth rate (1.8×10^{-5} in./cycle). The fatigue crack growth rate followed an approximate second power dependence ($m=2.05$). Growth properties were associated with a microstructure characterized by autotempered laths of predominantly dislocated martensite. The observed scatter in the crack growth rate data was ascribed to certain unidentified embrittling microstructural features, the presence of which was revealed by quasi-cleavage areas, as shown in Fig. 14.

The fracture toughness properties were very good. The fast fracture region, as shown in Fig. 16, showed primarily dimpled rupture appearance. This appearance is characteristic of high toughness materials.²² The high yield strength and fracture toughness properties of the steel in the as-quenched condition were attributed to the

relatively uniform microstructure, characterized by dislocated autotempered martensite. No residual carbides or extensive lath boundary precipitates were observed in the transmission studies. The yield strength of the as-quenched material was over 200,000 psi and the fracture toughness was about 90 ksi-in.^{1/2}.

b. 225°C tempered steel. Tempering at 225°C resulted in a significant increase in fracture toughness, as shown in Fig. 7, and a drop in yield strength, which was shown in Fig. 6. Transmission electron microscopy revealed that tempering at 225°C led to no obvious alteration in the microstructure from that in the as-quenched condition,¹⁷ although the lower yield strength reflected normal tempering effects, such as the growth of carbide particles.

The fatigue crack growth rates in the tempered samples were also characterized by a second power dependence (1.9); the growth rates for a given ΔK were essentially the same as those for the as-quenched specimens. The appearance of the fatigue fracture surface was the same as found in the as-quenched condition, as shown in Fig. 17. Extensive secondary cracking was observed in all fatigue regions as shown for the 225°C treatment. Figure 18 reveals the shallow-dimpled rupture appearance of the fast feature area, characteristic of the high fracture toughness.²² This, too, was similar to the appearance of the fracture surface of as-quenched specimens.

c. 300°C tempered steel. Tempering at 300°C led to a significant drop in yield strength, as well as a drop in fracture toughness, as shown in Figs 6 and 7. Transmission electron microscopy revealed that

substantial growth of lath boundary precipitates had occurred.¹⁷ The lath boundary precipitation led to a change in appearance in the fast fracture area. There was a change from dimpled rupture, characteristic of the 225°C tempered specimen, to quasi-cleavage. Figure 19 shows a scanning electron micrograph of a surface in the fast fracture region. Quasi-cleavage fracture features are evident in the figure. The fatigue crack growth rates did not reflect the same deterioration as the fracture toughness. They did, in fact, improve slightly. The fatigue fracture surface appearance was unchanged as shown in Fig. 20, and the coefficient, m , remained approximately 2 (1.9). There was no increase in growth rates even when maximum stress intensity, K_{max} , approached K_{IC} .

The deterioration of fracture toughness in the 260-315°C tempering range is usually referred to as "tempered martensite embrittlement" and is a common characteristic of quenched and tempered steels.²³ The embrittling effect is attributed to the preferential lath boundary precipitation of cementite.^{17,24,25} Boundary precipitates provide a crack path with a high density of brittle particles, which decreases the fracture strength. Fracture properties in quenched and tempered steels have been shown to be dependent on brittle particle spacing.²⁶

The structural features causing tempered martensite embrittlement were unexpectedly found to lead to an improvement in crack propagation rates in this steel. A similar discrepancy between toughness and crack growth rates was observed by Ancia and Kula in 4340 steel.¹²

Because the fatigue crack growth properties are not deleteriously affected by the presence of lath boundary precipitates it can be concluded that some of the microstructural features controlling fracture toughness do not affect the fatigue crack propagation in the same manner. Plastic deformation of the matrix is a requisite features of fatigue crack growth, and for a given set of test conditions, the characteristics of the matrix largely governs the crack growth rates. The influence of brittle particles is unclear; there is conflicting evidence on the role of such particles.²⁷⁻²⁹ Pelloux²⁸ found increased growth rates in aluminum alloys with increased particle density, but in his experiments the changes in particles density were large. Glassman and McEvily²⁹ found decreased rates in aluminum alloys with an increase in particle density. This work, however, was carried out at very high stresses with thin sheets and under conditions of plane stress.

Static fracture does not necessarily require plastic deformation; failure may be by cleavage in bcc alloys. Fracture is highly sensitive to microstructural features, especially those offering easy crack paths. Thus, it is not surprising that certain heat treating procedures may cause a severe drop in fracture toughness without appreciably changing fatigue crack growth behavior.

d. 500°C tempered steel. Tempering at 500°C led to a small increase in yield strength, as shown in Fig. 6. This increase was caused by secondary hardening. There was a concomitant drop in fracture toughness, as shown in Fig. 7. Transmission electron microscopy

revealed Fe_3C precipitate networks at lath boundaries.¹⁷ Although no molybdenum carbides were detected, they were undoubtedly present in small quantities because the 600°C tempering treatment (discussed in detail later) produced substantial amounts of easily identified molybdenum carbide.

The fracture crack growth rate increased considerably on tempering at 500°C. The exponent, m , was 2.55. Figure 21 shows the appearance of the fatigue region at ΔK of 31 ksi-in.^{1/2}. The corresponding crack growth rate was 2.5×10^{-5} in./cycle. The composite photograph reveals that the fatigue surface contained features similar to those observed in the 300°C temper condition. It was concluded that the thickened and extensive cementite films, as well as alloy carbides, led to the observed increase in growth rates for the 500°C tempered specimen. The fast fracture region, shown in Fig. 22, shows the same quasi-cleavage behavior as exhibited by the 300°C treatment, as would be expected.

e. 600°C tempered steel. A large increase in yield strength and a further decrease in fracture toughness were observed for specimens tempered at 600°C. The results are shown in Figs. 6 and 7. Transmission electron microscopy revealed Mo_2C dispersed within the martensite laths; electron diffraction showed that Fe_3C was no longer present in specimens tempered at this temperature.^{17,30} Extensive precipitation was revealed by optical metallography, as shown in Fig. 23.

The fatigue crack growth rates increased significantly over those characteristic of specimens tempered at lower temperatures. As K_{max} approached K_{IC} the growth rates increased to very high values, as shown in Fig. 12. The exponent, m , increased to 4.5, and there was

considerable scatter in the growth rate data. Scanning electron microscopy of the fatigue surface revealed the notable reasons for the increased crack growth rate and the cause of the larger scatter in the data. Figure 24 is a composite photograph of the fatigue surface taken from a specimen which exhibited a crack growth rate of 1×10^{-5} in./cycle at ΔK of 21.5 ksi-in.^{1/2}. The figure shows that an intergranular failure mode was present in the fatigue region, as well as the usual ductile mode characteristic of a fatigue fracture. In the low stress intensity ranges, the crack growth rate curves for specimens tempered at both 500°C and 600°C overlapped, suggesting a basic similarity in behavior in the lower stress range. The increase in growth rates and the large scatter for the 600°C tempered specimen can be attributed to the added intergranular mode of failure present during fatigue in the specimen tempered at 600°C. Similar observations have been made on other steels.⁶ Richie and Knott³¹ observed the presence of grain boundary cracking during fatigue in "temper brittle" EN30A (embrittled at 580°C) under plane strain conditions. Richie and Knott also found an increase in the exponent describing the fatigue crack growth properties. The embrittling phenomenon, leading to the intergranular cracking, has been attributed to several mechanisms. One is the segregation of impurity elements to prior austenite grain boundaries.³² Precipitation at prior austenite grain boundaries could also lead to deteriorated properties by providing a nearly continuous film.³³

The fast-fracture region is shown in Fig. 25. The characteristic quasi-cleavage appearance is shown as would be expected for low toughness behavior. Intergranular modes of failure are also observed as shown in Fig. 25, indicating the presence of weak films at prior austenite grain boundaries.

B. Results and Discussion of Treatments Subsequent to Austenization at 870°C

1. Mechanical Properties

a. Hardness, tensile, and fracture toughness. The 5Mo-.3C steel, although secondary hardening in nature, showed little in the way of properties associated with this category of materials. This is due to the low austenitization treatment temperature. The hardness vs tempering temperature curve is shown in Fig. 27. Because of the lack of carbide dissolution, the as-quenched hardness and subsequent hardness values are low. Although there are changes in behavior, the values are all very low. Only with tempering at 600°C is there some increase that might be connected with some small amount of alloy carbide formation. The tensile properties for the different treatments are listed in Table II. The yield strength variation, in particular, is displayed in Fig. 27, showing behavior that follows the hardness behavior.

The fracture toughness values are listed in Table II and shown in Fig. 28. The initial as-quenched fracture toughness is fairly low, as would be expected with the large volume fraction of brittle particles dispersed throughout the matrix. The values increase

somewhat with tempering. The only change in trend was seen for the 600°C tempering treatment, where the value decreased again to the level seen in the as-quenched condition.

b. Fatigue crack growth properties. The fatigue crack growth rates for different ΔK values were determined for all tempering treatments. The data are listed in Tables VII through XII. The Paris relation¹ was used for this steel after austenitization at 1200°C. The plots are shown in Figs. 29 through 33. The empirically determined coefficients are listed on the respective plots. Duplicate tests were conducted for all treatments except the as-quenched steel and the 225°C tempering treatment.

2. Microstructure and Crack Growth Behavior

Figure 34 is an optical photomicrograph showing the typical microstructure after quenching from the 870°C austenitizing temperature. Immediately apparent was the high volume fraction of carbide particles. These spheriodized carbides were of two size distributions; the larger were 1-3 μ in diameter while the smaller ones were 0.1 - 0.2 μ in diameter. This was seen earlier by Tom.¹⁸ He identified the carbides, using extraction replica, as being of the $(FeMo)_6C$ type. The surrounding microstructure was very fine due to the fine prior austenite grain size (ASTM 6). From a section through the ternary phase diagram, it can be seen that, at this low austenitizing temperature, the steel could easily lie in a three phase region.³⁴ Ferrite could still be retained at this temperature along with the austenite and carbide phases. The microstructure showed evidence of light areas, characteristic

of ferrite regions. The hardness values, as shown in Fig. 26, are low for a martensite-carbide material. The yield strength values, too, are low as suggested by Fig. 27. Isothermal transformation data at 480°C following austenitization at 870°C and rapid quenching to temperature, suggested too that approximately 5% ferrite could be present after processing. Because of the suspected presence of retained ferrite throughout the microstructure, it is not surprising that all mechanical properties, hardness and yield strength, remain fairly constant with tempering treatment, as shown in Figs. 26 and 27.

The fatigue crack growth properties also remained unchanged with the exception of the 600°C tempering treatment. The data for the other four treatments belonged to essentially one scatter band. The crack growth properties, too, were very good. The fatigue crack growth rates were much different from those observed for the higher austenitizing treatment. The empirical coefficient, m , describing the slope was larger; around 3. However, the initial rates of crack growth at low ΔK values ($\Delta K < 40 \text{ ksi-in}^{1/2}$) were much lower than those observed with the higher austenitization temperature. Only the 600°C tempering treatment was close in behavior in the ΔK range to those observed for the higher austenitization treatment.

Figure 35 is a composite scanning electron micrograph of the fatigue fracture surface for an as-quenched specimen. The region shows the fatigue region bordering the fast fracture region. Throughout the fatigue region there were evidences of many carbide particles dispersed through the matrix. These did occur randomly, however, and

did not appear to be the controlling microstructural feature for crack propagation. The fatigue surface did show diffuse striations, evidence of the complex microstructure. There were no areas of quasi-cleavage as found in the material austenitized at 1200°C. Although it was suspected that ferrite regions controlled crack growth, it was not apparent from the fracture surface.

The fast fracture surface is shown in Fig. 36. The surface showed characteristic quasi-cleavage appearance. However, the areas of cleavage were much smaller due to the smaller microstructural features in this steel. Carbide particles were clearly seen on the fast fracture surface also.

Although the behavior in the material tempered at 600°C differed, the same problems existed in attempting to distinguish the controlling microstructural features from the fatigue crack surface. The deterioration in crack growth properties could be attributed to the slight secondary hardening that might be taking place causing an upswing in hardness and deterioration in fracture toughness.

With the lower austenitization treatment, three microstructural changes occurred. (1) Carbide particles in large volume were present. (2) Prior austenization grain size was small leading to smaller as-quenched microstructural features. (3) Ferrite retained during austenitization could be present. All these features could influence fatigue crack growth rates. However, it was impossible to judge which did help improve properties, from the fatigue fracture surfaces.

C. Results and Discussion of Additional Work

In an attempt to explain the excellent fatigue crack growth properties found for the 5Mo-.3C steel after austenitization at 870°C, fatigue crack growth rates for another austenitizing treatment and some additional compositions were investigated. Austenitization at 980°C for the 5Mo-.3C steel was undertaken to attempt to decide whether the carbide particles or ferrite led to the fatigue property improvement. Two compositions of lower Mo content were investigated to determine whether prior austenite grain size led to the improvement.

1. Austenitization of 5Mo-.3C at 980°C

Austenitization at 980°C for one hour, 110°C higher than that producing improved crack growth rates, followed by an ice-brine quench, was investigated. The microstructure produced contained a high volume fraction of carbides, as shown in Fig. 37. The hardness, of the as-quenched material, H_v 490, was definitely higher than that of the 870°C treatment, indicating an absence of any ferrite during austenitization.

The fatigue crack growth properties are listed in Table XIII and shown in Fig. 38. The crack growth rates are again characterized by a coefficient close to 2(1.7) and lie in the same scatter band as those for the 1200°C austenitized, as-quenched material. The rates are worse at low ΔK values than those for the 870°C austenitization treatments. The fatigue fracture surface as shown by SEM in Fig. 39

contains features similar to surfaces for the 1200°C treatment. The fracture region is shown in Fig. 40. It contained quasi-cleavage features similar to the 870°C austenitized material, and exhibited similar fracture behavior. From this test, it could be concluded that the large carbide particles were not the feature that improved properties.

2. Investigation of 1Mo-.24C, 1Mo-.34C Steel After Austenitization at 870°C

Investigation of these lower Mo-steels was undertaken to clarify the role of prior austenite grain size and the martensite morphology it produced on fatigue crack propagation. The microstructures of the 1Mo-.24C, 1Mo-.34C steels after quenching are shown in Figs. 41 and 42, respectively. Both show a martensitic structure similar to that of the 5Mo-.3C steel after austenitization at 1200°C. The hardness values, H_v 542 and H_v 624, respectively indicate the martensitic nature of the steels. Tom in his study used replicas to show that the alloys had no carbides present.¹⁸

The fatigue crack growth properties are listed in Tables XIV and XV. Figures 43 and 44 show the rates correlated with ΔK levels using the Paris relation.¹ The empirically determined slopes, are around 2.5 for both. The fatigue fracture surfaces are shown in Figs. 45 and 46. The appearance is basically the same for both alloys, and very similar to the surfaces for the 1200°C austenitized 5Mo-.3C steel. Finally the fast fracture surfaces, shown in Figs. 47 and 48, both showed dimpled rupture appearance, indicating the high fracture toughness values as seen by Tom for these alloys.¹⁸

Although the martensite structure size was smaller from that for the 5Mo-.3C material after high austenitization, there seems to be no apparent change in fatigue crack growth rates. Martensite morphology, too, may have changed. Kelly and Nutting²¹ observed a change from dislocated martensite to twinned martensite as carbon level increased. Therefore the smaller microstructural features brought about by lower austenitization temperatures were not responsible for the improved fatigue crack growth properties observed in the 5Mo-.3C steel.

IV. SUMMARY AND CONCLUSIONS

The fatigue crack growth properties and fracture toughness properties for the high austenitized 5Mo-0.3C steel in the as-quenched condition are compared with the other tempering treatments in Fig. 49 (the crack growth properties are shown as the number of cycles to grow one inch at a given ΔK). This figure clearly shows that there is no unique relation between fatigue crack growth rates and fracture toughness. Specimens tempered at 300°C had lower values of fracture toughness, whereas they exhibited a reduction in crack growth rate when compared with specimens tempered at 225°C.

Figure 50 shows the fatigue crack propagation rates of the as-quenched 5Mo-0.3C steel austenitized at 1200°C along with crack growth rates for martensitic steels having yield strengths of about 200,000 psi.^{9,10,12,35} As the figure shows, the as-quenched 5Mo-0.3C steel displayed fatigue crack propagation rates at high values of ΔK that were appreciably less than those reported for other steels.

Comparing the 870°C austenitizing treatment with the 1200°C treatment, different conclusions are reached, however. Figure 51 shows the fatigue crack growth rates compared for the two austenitization treatments. It is easily seen that the low austenitizing treatment had much better properties. Although the fracture toughness is lower for the 870°C austenitized material, the fatigue crack growth properties control the lifetime for this steel when it is used under cyclic loading. Figure 52 compares the cycles to failure for the two treatments for

different loads, given a specific flaw size and configuration. For all cases except very high loads, the low austenitization treatments led to longer lifetimes even though a_c was smaller.

The general conclusions gleaned from the study of fatigue crack propagation in 5Mo-0.3C steel are summarized below:

(1) Microstructural changes in quenched steels brought about by austenitization or tempering treatments can lead to drastic changes in fatigue crack growth rates.

(2) The microstructural features affecting fracture toughness do not always affect fatigue crack propagation in the same way. There is no unique relationship between the fracture toughness and crack growth rates that holds for all microstructures. In the case of 5Mo-.3C steel austenitized at 1200°C, cementite precipitation at lath boundaries lowered fracture toughness while having no deleterious effect on the fatigue crack growth rates. When the carbide precipitation became more extensive, however, both properties deteriorated. In the case of the 5Mo-.3C steel austenitized at 870°C, a microstructure containing a high volume fraction of carbides had excellent fatigue crack growth properties and poor fracture toughness.

(3) The presence of additional failure modes in the fatigue area, such as intergranular fracture, causes an increase in crack growth rates, as characterized by the empirical constants in the Paris equation.

(4) Fatigue crack propagation can overweigh fracture toughness in importance in determining a steel's usefulness for certain applications.

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Table I. Chemical analyses of alloys used.

Alloy No.	Mo	Mn	C	Fe
723-15	5.0	0.66	0.28	balance
723-16	5.1	0.70	0.26	balance
723-18	5.2	0.62	0.28	balance
321	1.0	.66	0.24	balance
324	1.0	.66	0.34	balance

Table II. Mechanical Properties of 5Mo-.3C Steel.

A. Austenitized @ 1200°C					
Tempering Treatment	.2% Y.S.	UTS	Elong. %	R. A. %	Fracture Toughness (a/W = .7)
As-Quenched	217	274	8.3	20.4	92 ² ksi-in ^{1/2}
225°C	202	239	10.6	29.6	103 ²
300°C	179	223	10.0	33.7	76.5
500°C	187	227	10.5	33.8	45 ²
600°C ¹	251	259	1.0	5	36 ²
B. Austenitized @ 870°C					
Tempering Treatment	.2% Y.S.	UTS	Elong. %	R. A. %	Fracture Toughness (a/W = .7)
As-Quenched	182	216	11.0	46	67
225°C	186	213	11.6	36.8	70
300°C	164	185	13.1	53.5	66 ²
500°C	171	178	17.3	59	83 ²
600°C	171	175	16.5	50.4	64 ²

1 - Ref. 17

2 - avg. values

Table III. Fatigue Crack Growth Data-2200°F Aust.; As-Quenched.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
15-1	5000	.828	.861	300	2700	23.40
	2000	.861	.887	300	2700	24.35
	2000	.887	.905	300	2700	25.00
	2000	.910*	.933	300	2700	25.69
	2000	.933	.954	300	2700	26.58
	1000	.954	.960	300	2700	27.15
	2000	.960	.985	300	2700	27.75
	2000	.985	1.008	300	2700	28.80
	2000	1.008	1.034	300	2700	30.11
	2000	1.034	1.058	300	2700	31.30
	1000	1.058	1.072	300	2700	32.06
	1000	1.072	1.085	300	2700	32.60
	1000	1.082*	1.096	300	2700	33.42
	1000	1.096	1.111	300	2700	34.28
	1000	1.111	1.127	300	2700	35.20
	1000	1.127	1.142	300	2700	36.15
	1000	1.142	1.159	300	2700	37.15
	1000	1.159	1.171	300	2700	38.20
	1000	1.171	1.186	300	2700	39.32
	1000	1.186	1.206	300	2700	40.38
	1000	1.206	1.223	300	2700	40.82
	1000	1.223	1.242	300	2700	43.19
	1000	1.242	1.268	300	2700	45.27
	500	1.268	1.283	300	2700	47.58
	500	1.283	1.294	300	2700	48.04
	500	1.294	1.309	300	2700	50.35
	500	1.309	1.320	300	2700	51.48
	500	1.320	1.334	300	2700	53.72
	1000	1.334	1.374	300	2700	56.03
	500	1.374	1.395	300	2700	60.33
18-1	10,000	1.092	1.132	300	1950	23.07
	10,000	1.132	1.170	300	1780	22.93
	10,000	1.170	1.204	600	1780	19.38
	10,000	1.204	1.231	600	1680	18.87
	10,000	1.231	1.270	600	1680	20.13

* Specimen removed and crack lengths measured using a metallurgical microscope with a calibrated stage.

Table IV. Fatigue Crack Growth Data; 2200°F Aust; Tempered at 225°C.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max Load (lb.)	ΔK -ksi-in. ^{1/2}
18-2	5000	.831	.870	300	2700	23.56
	5000	.870	.906	300	2700	24.66
	5000	.906	.946	300	2700	25.87
	4000	.946	.979	300	2700	27.36
	2000	.979	.991	300	2700	28.35
	2000	.991	1.006	300	2700	28.71
	2000	1.006	1.019	300	2670	28.92
	2000	1.019	1.039	300	2700	29.96
	2000	1.039	1.056	300	2670	30.66
	2000	1.056	1.077	300	2700	32.08
	2000	1.077	1.101	300	2670	33.00
	1000	1.101	1.110	300	2700	34.07
	1000	1.110	1.124	300	2700	34.61
	1000	1.124	1.133	300	2670	34.79
	1000	1.133	1.144	300	2640	34.91
	1000	1.144	1.159	300	2610	35.66
	1000	1.159	1.172	300	2700	38.20
	1000	1.172	1.187	300	2700	39.07
	1000	1.187	1.194	300	2670	39.71
	1000	1.194	1.214	300	2640	40.00
	1000	1.214	1.228	300	2610	40.96
	1000	1.228	1.248	300	2700	43.82
	1000	1.248	1.270	300	2640	45.17
	1000	1.270	1.293	300	2700	48.43
	500	1.293	1.306	300	2700	50.08
	500	1.306	1.315	300	2640	49.96
	500	1.315	1.331	300	2700	52.77
	500	1.331	1.346	300	2670	54.65
	500	1.346	1.363	300	2700	57.03
	500	1.363	1.379	300	2640	58.47
	500	1.379	1.398	300	2700	63.09
16-2	2000	.985	1.003	300	2700	25.66
	2000	1.003	1.023	300	2700	26.38
	2000	1.023	1.038	300	2700	27.20
	2000	1.038	1.061	300	2700	28.05
	2000	1.061	1.081	300	2700	28.96
	2000	1.081	1.104	300	2700	30.18
	2000	1.104	1.121	300	2700	31.20
	2000	1.121	1.146	300	2700	32.12
	1000	1.146	1.156	300	2700	33.28
	1000	1.156	1.164	300	2700	33.80

Table IV. (Cont.)

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
16-2	1000	1.164	1.177	300	2700	34.23
	1000	1.177	1.188	300	2700	35.21
	1000	1.188	1.204	300	2700	35.90
	1000	1.204	1.220	300	2700	36.98
	1000	1.220	1.233	300	2700	37.86
	1000	1.233	1.250	300	2700	39.06
	1000	1.250	1.266	300	2700	40.48
	1000	1.266	1.284	300	2700	41.82
	1000	1.284	1.305	300	2700	43.74
	1000	1.305	1.328	300	2700	46.00
	1000	1.328	1.350	300	2700	48.23
	1000	1.350	1.384	300	2700	51.26
	1000	1.384	1.399	300	2700	54.54

Table V. Fatigue Crack Data; Aust. @ 1200°C, Tempered @ 300°C.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
15-3	2000	.839	.844	300	2700	23.48
	2000	.844	.848	300	2700	23.63
	5000	.848	.872	300	2700	24.10
	5000	.872	.906	300	2700	24.90
	2000	.906	.917	300	2700	25.75
	2000	.917	.928	300	2700	26.29
	2000	.928	.934	300	2700	26.47
	2000	.934	.948	300	2700	26.85
	2000	.948	.962	300	2700	27.41
	2000	.962	.973	300	2700	27.81
	2000	.973	.989	300	2700	28.43
	2000	.989	1.008	300	2700	29.27
	2000	1.008	1.025	300	2700	29.94
	2000	1.025	1.053	300	2700	30.89
	2000	1.053	1.070	300	2700	32.13
	2000	1.070	1.089	300	2700	33.20
	2000	1.089	1.114	300	2700	34.34
	1000	1.114	1.126	300	2700	35.56
	1000	1.126	1.137	300	2700	36.50
	1000	1.137	1.149	300	2700	37.17
	1000	1.149	1.160	300	2700	37.86
	1000	1.160	1.174	300	2700	38.57
	1000	1.174	1.188	300	2700	39.71
	1000	1.188	1.207	300	2700	41.30
	1000	1.207	1.222	300	2700	42.55
	1000	1.222	1.236	300	2700	43.90
	1000	1.236	1.255	300	2700	45.79
	1000	1.255	1.276	300	2700	47.00
	1000	1.276	1.293	300	2700	48.28
	1000	1.293	1.319	300	2700	51.13
	1000	1.319	1.339	300	2700	54.56
	1000	1.339	1.369	300	2700	57.24
	1000	1.369	1.399	300	2700	62.12
16-3	5000	.941	.953	300	2700	24.20
	2000	.953	.970	300	2700	24.71
	2000	.970	.982	300	2700	25.06
	2000	.982	1.001	300	2700	25.81
	2000	1.001	1.018	300	2700	26.58
	2000	1.018	1.030	300	2700	26.98
	2000	1.030	1.051	300	2700	27.44
	2000	1.051	1.068	300	2700	28.32
	2000	1.068	1.085	300	2700	29.06
	2000	1.085	1.106	300	2700	29.84

Table V. (Cont.)

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
16-3	2000	1.106	1.127	300	2700	30.88
	2000	1.127	1.148	300	2700	31.94
	1000	1.148	1.162	300	2700	33.08
	1000	1.162	1.173	300	2700	33.68
	1000	1.173	1.186	300	2700	34.66
	1000	1.186	1.200	300	2700	35.68
	1000	1.200	1.214	300	2700	36.38
	1000	1.214	1.227	300	2700	37.50
	1000	1.227	1.241	300	2700	38.70
	1000	1.241	1.259	300	2700	39.63
	1000	1.259	1.281	300	2700	41.63
	1000	1.281	1.299	300	2700	43.24
	1000	1.299	1.316	300	2700	44.72
	1000	1.316	1.337	300	2700	46.86
	1000	1.337	1.360	300	2700	49.40
	500	1.360	1.379	300	2700	51.88

Table VI. Fatigue Crack Growth Data; Aust. @1200°C; Tempered @500°C.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
18-4	5000	.834	.868	300	2700	17.85
	2000	.868	.885	300	2700	18.43
	2000	.885	.903	300	2700	18.80
	2000	.903	.916	300	2700	19.32
	2000	.916	.932	300	2700	19.72
	2000	.932	.952	300	2700	19.96
	2000	.952	.971	300	2700	20.70
	2000	.971	.993	300	2700	21.31
	2000	.993	1.011	300	2700	21.96
	2000	1.011	1.036	300	2700	22.81
	2000	1.036	1.056	300	2700	23.52
	2000	1.056	1.084	300	2700	24.30
	2000	1.084	1.110	300	2700	25.10
	2000	1.110	1.148	300	2700	26.70
	2000	1.148	1.183	300	2700	28.45
	1000	1.183	1.200	300	2700	29.85
	1000	1.200	1.225	300	2700	31.06
	1000	1.225	1.251	300	2700	32.72
	1000	1.251	1.277	300	2700	34.90
	1000	1.277	1.326	300	2700	37.55
	1000	1.326	1.348	300	2400	35.16
	1000	1.348	1.375	300	2400	36.33
15-4	10,000	1.130	1.151	600	1560	14.36
	10,000	1.151	1.170	600	1560	15.14
	10,000	1.170	1.200	600	1560	15.63
	10,000	1.200	1.236	600	1560	16.78
	10,000	1.236	1.286	600	1560	17.98

Table VII. Fatigue Crack Growth Data; Aust. @1200°C; Tempered @600°C.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
15-5	2000	.919	.923	180	1500	14.71
	2000	.923	.932	180	1500	14.91
	2000	.932	.935	180	1500	15.01
	2000	.935	.947	180	1500	15.12
	2000	.947	.952	180	1500	15.33
	4000	.952	.958	180	1500	15.44
	4000	.958	.970	180	1500	15.55
	5000	.970	.983	180	1500	15.89
	5000	.983	1.005	180	1500	16.37
	5000	1.005	1.025	180	1500	16.87
	2000	1.025	1.044	180	1500	18.98
	2000	1.044	1.054	180	1620	19.40
	2000	1.054	1.070	180	1620	19.72
	2000	1.070	1.084	180	1620	20.21
	2000	1.084	1.098	180	1620	20.71
	2000	1.098	1.115	180	1620	21.03
	2000	1.115	1.153	180	1620	21.96
	1000	1.153	1.183	180	1620	23.19
	1000	1.183	1.236	180	1620	24.80
	100	1.236	1.248	180	1620	26.88
	100	1.248	1.255	180	1500	25.70
	100	1.255	1.261	180	1500	25.65
	100	1.261	1.278	180	1500	26.22
	1000	1.279	1.289	180	1200	21.20
	1000	1.289	1.309	180	1200	21.92
	1000	1.309	1.331	180	1200	22.92
	1000	1.331	1.356	180	1200	23.35

Table VIII. Fatigue Crack Growth Data; Aust. @ 870°C; As-Quenched.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
18-6-2	5000	.921	.930	300	2700	26.49
	5000	.930	.941	300	2700	26.66
	5000	.941	.960	300	2700	27.22
	5000	.960	.982	300	2700	28.01
	5000	.982	1.000	300	2700	28.84
	5000	.997*	1.018	300	2700	29.72
	5000	1.018	1.043	300	2700	30.64
	5000	1.043	1.069	300	2700	31.88
	2000	1.069	1.082	300	2700	32.92
	5000	1.082	1.118	300	2700	34.36
	2000	1.118	1.134	300	2700	35.86
	2000	1.134	1.151	300	2700	36.84
	2000	1.151	1.171	300	2700	38.22
	2000	1.171	1.194	300	2700	39.46
	2500	1.194	1.227	300	2700	41.90
	1000	1.227	1.241	300	2700	44.06
	1000	1.241	1.252	300	2700	45.01
	1000	1.252	1.276	300	2700	47.03
	1000	1.276	1.293	300	2700	48.83
	1000	1.293	1.315	300	2700	50.20
	500	1.315	1.327	300	2700	52.94
		1.327	1.359	300	2700	55.50
		1.359	1.400	300	2700	60.56

* Specimen removed and crack lengths measured using a metallurgical microscope with a calibrated stage.

Table IX. Fatigue Crack Growth Data; Aust. @ 870°C; Tempered @ 225°C

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
18-6-3	5000	.837	.843	300	2700	23.33
	5000	.843	.857	300	2700	23.63
	5000	.857	.868	300	2700	23.94
	2500	.868	.884	300	3000	27.65
	5000	.884	.916	300	3000	28.40
	2000	.916	.926	300	3000	29.18
	2000	.926	.939	300	3000	29.77
	2000	.939	.954	300	3000	30.41
	5000	.954	.968	300	2700	27.41
	5000	.968	.985	300	2700	28.01
	5000	.985	1.010	300	2700	29.06
	5000	1.010	1.036	300	2700	30.17
	5000	1.036	1.063	300	2700	31.37
	5000	1.063	1.101	300	2700	33.22
	2000	1.101	1.117	300	2700	34.62
	2000	1.117	1.136	300	2700	35.54
	2000	1.136	1.153	300	2700	36.84
	2000	1.153	1.173	300	2700	38.22
	2000	1.173	1.193	300	2700	39.70
	2000	1.193	1.222	300	2700	41.73
	1000	1.222	1.237	300	2700	43.90
	1000	1.237	1.254	300	2700	45.30
	1000	1.254	1.274	300	2700	46.20
	1000	1.274	1.296	300	2700	48.85
	1000	1.296	1.321	300	2700	51.71
	1000	1.321	1.346	300	2700	53.85
	1000	1.346	1.379	300	2700	57.60

Table X. Fatigue Crack Growth Data; Aust. @ 870°C; Tempered @ 300°C.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
15-6-5	5000	.854	.866	300	2700	24.25
	5000	.866	.882	300	2700	24.58
	5000	.882	.896	300	2700	25.07
	5000	.896	.906	300	2700	25.58
	5000	.906	.924	300	2700	25.93
	5000	.924	.943	300	2700	26.48
	5000	.943	.966	300	2700	27.42
	5000	.966	.973	300	2700	28.01
	5000	.973	.990	300	2700	28.43
	5000	.990	1.007	300	2700	29.27
	5000	1.007	1.029	300	2700	30.17
	5000	1.029	1.051	300	2700	31.12
	5000	1.051	1.076	300	2700	32.40
	5000	1.076	1.104	300	2700	33.54
	5000	1.104	1.133	300	2700	35.02
	2000	1.133	1.149	300	2700	36.84
	2000	1.149	1.166	300	2700	38.23
	2000	1.182	1.199	300	2700	40.49
	2000	1.199	1.216	300	2700	42.13
	2000	1.216	1.236	300	2700	43.62
	2000	1.236	1.258	300	2700	45.50
	2000	1.258	1.286	300	2700	48.08
	2000	1.286	1.312	300	2700	50.53
	2000	1.312	1.348	300	2700	53.90
	1000	1.348	1.369	300	2700	58.32
18-6-6	5000	.859	.866	300	2700	23.86
	5000	.866	.874	300	2700	24.02
	5000	.874	.886	300	2700	24.50
	5000	.886	.899	300	2700	25.00
	5000	.900*	.904	300	2700	25.16
	5000	.904	.915	300	2700	25.33
	5000	.915	.931	300	2700	25.86
	5000	.931	.947	300	2700	26.58
	5000	.947	.963	300	2700	27.14
	5000	.963	.980	300	2700	27.75

Table X. (Cont.)

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
18-6-6	5000	.980	1.005	300	2700	28.74
	5000	1.005	1.029	300	2700	29.83
	5000	1.029	1.056	300	2700	31.07
	5000	1.056	1.088	300	2700	32.61
	2000	1.088	1.103	300	2700	33.68
	2000	1.103	1.117	300	2700	34.57
	2000	1.117	1.134	300	2700	35.50
	2000	1.134	1.151	300	2700	36.48
	2000	1.151	1.170	300	2700	37.86
	2000	1.170	1.192	300	2700	39.07
	2000	1.192	1.214	300	2700	40.64
	2000	1.214	1.238	300	2700	42.48
	2000	1.238	1.271	300	2700	44.97
	2000	1.271	1.312	300	2700	48.30
	1000	1.312	1.340	300	2700	52.72
		1.340	1.362	300	2700	55.98

* Specimen removed and crack lengths measured using a metallurgical microscope with a calibrated stage.

Table XI. Fatigue Crack Growth Data; Aust. @ 870°C; Tempered @ 500°C.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
18-6-7	5000	.806	.817	300	2700	22.51
	5000	.817	.830	300	2700	22.80
	5000	.830	.837	300	2700	23.10
	5000	.837	.856	300	2700	23.55
	5000	.856	.872	300	2700	24.02
	5000	.872	.889	300	2700	24.51
	5000	.889	.905	300	2700	25.00
	5000	.905	.922	300	2700	25.50
	5000	.922	.946	300	2700	26.40
	5000	.946	.969	300	2700	27.34
	5000	.969	.993	300	2700	28.15
	5000	.993	1.020	300	2700	29.21
	5000	1.020	1.050	300	2700	30.59
	2000	1.050	1.059	300	2700	31.55
	2000	1.059	1.071	300	2700	32.06
	2000	1.071	1.083	300	2700	32.39
	2000	1.083	1.099	300	2700	32.96
	2000	1.099	1.113	300	2700	34.07
	2000	1.113	1.127	300	2700	34.98
	2000	1.127	1.143	300	2700	35.91
	2000	1.143	1.160	300	2700	36.92
	2000	1.160	1.179	300	2700	38.08
	2000	1.179	1.203	300	2700	39.59
	2000	1.203	1.258	300	2700	42.09
	2000	1.258	1.255	300	2700	43.82
	2000	1.255	1.286	300	2700	46.49
	1000	1.286	1.304	300	2700	49.45
	1000	1.304	1.324	300	2700	50.86
	1000	1.324	1.353	300	2700	53.99
	1000	1.353	1.381	300	2700	57.73
15-6-8	5000	.858	.860	300	2700	23.86
	5000	.860	.893	300	3000	27.39
	5000	.893	.911	300	2700	25.19
	5000	.911	.924	300	2700	25.67
	5000	.924	.944	300	2700	26.21

Table XI. (Cont.)

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
15-6-8	5000	.944	.969	300	2700	27.15
	5000	.969	.993	300	2700	28.14
	5000	.993	1.007	300	2700	28.76
	5000	1.007	1.047	300	2700	30.14
	2000	1.047	1.059	300	2700	31.55
	2000	1.059	1.069	300	2700	32.06
	2000	1.069	1.091	300	2700	32.88
	2000	1.091	1.103	300	2700	33.70
	2000	1.103	1.122	300	2700	34.59
	2000	1.122	1.136	300	2700	35.60
	2000	1.136	1.151	300	2700	36.58
	2000	1.151	1.176	300	2700	37.98
	2000	1.176	1.202	300	2700	39.58
	2000	1.202	1.226	300	2700	41.62
	2000	1.226	1.254	300	2700	43.54
	2000	1.254	1.300	300	2700	46.73
	1000	1.300	1.327	300	2700	51.49
	1000	1.327	1.364	300	2700	54.95

Table XII. Fatigue Crack Growth Data; Aust. @ 870°C; Tempered @ 600°C.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
15-6-9	5000	.837	.860	300	2700	23.79
	5000	.860	.883	300	2700	24.42
	5000	.883	.910	300	2700	25.25
	5000	.910	.943	300	2700	26.30
	5000	.943	.973	300	2700	27.43
	5000	.973	1.013	300	2700	28.06
	2000	1.013	1.031	300	2700	30.17
	2000	1.031	1.052	300	2700	31.12
	2000	1.052	1.069	300	2700	32.18
	2000	1.069	1.090	300	2700	33.20
	2000	1.090	1.116	300	2700	34.42
	2000	1.116	1.143	300	2700	35.85
	2000	1.143	1.171	300	2700	38.01
	2000	1.171	1.205	300	2700	40.24
	2000	1.205	1.219	300	2400	37.23
	1000	1.219	1.232	300	2400	38.00
	1000	1.232	1.247	300	2400	38.52
	1000	1.247	1.266	300	2400	39.93
	1000	1.266	1.287	300	2400	42.16
	1000	1.287	1.310	300	2400	43.80
	1000	1.310	1.336	300	2400	46.10
	1000	1.336	1.369	300	2400	49.56
	1000	1.369	1.391	300	2400	52.76
18-6-0	5000	.836	.848	300	2700	23.63
	5000	.848	.878	300	2700	24.26
	5000	.878	.890	300	2700	24.91
	5000	.890	.921	300	2700	25.59
	5000	.921	.956	300	2700	26.86
	3000	.956	.980	300	2700	28.02
	2000	.980	.998	300	2700	28.85
	2000	.998	1.016	300	2700	29.72
	2000	1.016	1.036	300	2700	30.64
	2000	1.036	1.058	300	2700	31.62
	2000	1.058	1.080	300	2700	32.45
	2000	1.080	1.105	300	2700	33.55

Table XII. (Cont.)

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
18-6-0	2000	1.105	1.135	300	2700	35.35
	1000	1.135	1.151	300	2700	36.94
	1000	1.151	1.169	300	2700	37.98
	1000	1.169	1.187	300	2700	38.82
	1000	1.187	1.209	300	2700	40.38
	1000	1.209	1.233	300	2700	42.18
	1000	1.233	1.261	300	2700	44.16
	500	1.261	1.284	300	2700	46.92
	500	1.284	1.303	300	2700	49.07
	500	1.303	1.326	300	2700	51.65
	500	1.326	1.352	300	2700	54.81

Table XIII. Fatigue Crack Growth Data; Aust. @ 980°C; As-Quenched.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
1800-1	5000	.900	.952	300	2700	26.77
	2000	.952	.976	300	2700	28.30
	2000	.976	1.000	300	2700	29.36
	2000	1.000	1.026	300	2700	30.46
	2000	1.026	1.054	300	2700	31.65
	2000	1.054	1.080	300	2700	33.25
	1000	1.080	1.096	300	2700	34.52
	1000	1.096	1.111	300	2700	35.42
	1000	1.111	1.128	300	2700	36.34
	1000	1.128	1.146	300	2700	37.08
	1000	1.146	1.163	300	2700	38.46
	1000	1.163	1.181	300	2700	39.92
	1000	1.181	1.203	300	2700	41.50
	1000	1.203	1.223	300	2700	43.18
	1000	1.223	1.246	300	2700	44.99
	1000	1.246	1.279	300	2700	47.16
	1000	1.279	1.313	300	2700	50.65
	500	1.313	1.338	300	2700	53.90
	100	1.338	1.347	300	2700	57.30
	100	1.347	1.355	300	2700	58.70
	255	1.355	1.371	300	2700	60.16

Table XIV. Fatigue Crack Growth Data; Aust. @ 870°C; As-Quenched.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
321	2000	.905	.910	300	2700	24.15
	2000	.910	.916	300	2700	24.32
	2000	.916	.924	300	2700	24.48
	2000	.924	.939	300	2700	24.82
	2000	.939	.957	300	2700	25.53
	2000	.957	.979	300	2700	26.27
	2000	.979	1.000	300	2700	27.05
	2000	1.000	1.022	300	2700	27.86
	2000	1.022	1.042	300	2700	28.72
	2000	1.042	1.064	300	2700	29.64
	1000	1.064	1.076	300	2700	30.62
	1000	1.076	1.087	300	2700	31.38
	1000	1.087	1.099	300	2700	31.92
	1000	1.099	1.110	300	2700	32.47
	1000	1.110	1.123	300	2700	33.04
	1000	1.123	1.138	300	2700	33.92
	1000	1.138	1.151	300	2700	34.85
	1000	1.151	1.169	300	2700	35.50
	1000	1.169	1.180	300	2700	36.52
	1000	1.188*	1.207	300	2700	38.72
	1000	1.207	1.222	300	2700	39.90
	1000	1.222	1.240	300	2700	40.90
	1000	1.240	1.262	300	2700	42.40
	1000	1.262	1.284	300	2700	44.00
	1000	1.284	1.309	300	2700	46.54
	1000	1.309	1.340	300	2700	50.57
	500	1.340	1.358	300	2700	53.35
	100	1.358	1.363	300	2700	55.02
	100	1.363	1.368	300	2700	55.37
	100	1.368	1.377	300	2700	63.46

* Specimen removed and crack lengths measured using a metallurgical microscope with a calibrated stage.

Table XV. Fatigue Crack Growth Data; Aust. @ 870°C; As-Quenched.

Specimen	Cycles	a_i (in.)	a_f (in.)	Min. Load (lb.)	Max. Load (lb.)	ΔK -ksi-in. ^{1/2}
324	2000	.971	.998	300	2700	27.27
	1000	.998	1.012	300	2700	28.10
	1000	1.012	1.021	300	2700	28.55
	1000	1.021	1.032	300	2700	28.94
	1000	1.032	1.047	300	2700	29.63
	1000	1.047	1.060	300	2700	30.34
	1000	1.060	1.072	300	2700	30.84
	1000	1.072	1.084	300	2700	31.34
	1000	1.084	1.100	300	2700	32.14
	1000	1.100	1.117	300	2700	33.25
	1000	1.117	1.133	300	2700	34.13
	1000	1.133	1.149	300	2700	35.08
	1000	1.149	1.165	300	2700	36.01
	1000	1.165	1.183	300	2700	37.09
	1000	1.183	1.200	300	2700	38.56
	1000	1.206*	1.224	300	2700	40.51
	1000	1.224	1.250	300	2700	42.25
	1000	1.250	1.269	300	2700	43.98
	1000	1.269	1.293	300	2700	45.98
	500	1.293	1.308	300	2700	48.42
	500	1.308	1.326	300	2700	50.40
	500	1.326	1.343	300	2700	52.23
	100	1.343	1.348	300	2700	53.85
	100	1.348	1.353	300	2700	54.50
	100	1.353	1.361	330	3000	61.79
	100	1.361	1.370	330	3000	62.98
	100	1.370	1.380	330	3000	64.20
	100	1.380	1.391	330	3000	65.83

* Specimen removed and crack lengths measured using a metallurgical microscope with a calibrated stage.

FIGURE CAPTIONS

- Fig. 1. Round tensile specimen.
- Fig. 2. Compact tension fracture toughness specimen; 9/16 in. thickness used.
- Fig. 3. Setup used for fatigue crack growth studies showing MTS and alignment scope.
- Fig. 4. Calibration curve for COD gage.
- Fig. 5. Hardness as a function of tempering temperature. 5Mo-.3C steel austenitized at 1200°C.
- Fig. 6. Yield strength as a function of tempering temperature 5Mo-.3C steel austenitized at 1200°C.
- Fig. 7. Plane strain fracture toughness as a function of tempering temperature. 5Mo-.3C; austenitized at 1200°C
- Fig. 8. Fatigue crack propagation rate vs stress intensity range for the as-quenched 5Mo-0.3C steel austenitized at 1200°C. Data from specimens 15-1 and 18-1. (Paris relation fitted for stress intensity in $\text{psi-in}^{1/2}$ units).
- Fig. 9. Fatigue crack propagation rates vs stress intensity range for 1200°C austenitized, 225°C tempered 5Mo-.3C steel. Data from specimens 16-2 and 18-2.
- Fig. 10. Fatigue crack propagation rates vs stress intensity range for 1200°C austenitized, 300°C tempered 5Mo-.3C steel. Data from specimens 16-2 and 18-3.

- Fig. 11. Fatigue crack propagation rates vs stress intensity range for 1200°C austenitized, 500°C tempered 5Mo-.3C steel. Data from specimens 15-4 and 18-4.
- Fig. 12. Fatigue crack propagation rates vs stress intensity range for 1200°C austenitized, 600°C tempered 5Mo-.3C steel. Data from specimen 15-5.
- Fig. 13. Microstructure of as-quenched 5Mo-.3C steel after austenitization at 1200°C.
- Fig. 14. Fatigue fracture region in 1200°C austenitized, as-quenched 5Mo-.3C steel. (a) Overall appearance of surface. $\Delta K=35 \text{ ksi-in.}^{1/2}$. (b) Composite enlargement showing quasi-cleavage at A, fatigue cracking through bulk microstructure at B, and local striations at C.
- Fig. 15. Fatigue fracture showing striations in 1200°C austenitized as-quenched 5Mo-.3C steel. (a) Overall appearance of fatigue surface. $\Delta K=40 \text{ ksi-in.}^{1/2}$. Striations appear at A. (b) Enlargement of striations.
- Fig. 16. Fast fracture region of 1200°C austenitized, as-quenched 5Mo-.3C steel.
- Fig. 17. Fatigue fracture surface of 1200°C austenitized, 225°C tempered 5Mo-.3C steel.
- Fig. 18. Fast fracture appearance of 1200°C austenitized, 225°C tempered 5Mo-.3C steel showing primarily dimpled rupture appearance.

- Fig. 19. Fast fracture appearance of 1200°C austenitized, 300°C tempered 5Mo-.3C steel showing primarily quasi-cleavage appearance.
- Fig. 20. Composite of fatigue fracture surface of 1200°C austenitized 300°C tempered 5Mo-.3C steel. General fatigue appearance resembles that of as-quenched material. $\Delta K=35 \text{ ksi-in.}^{1/2}$.
- Fig. 21. Composite SEM fractograph showing fatigue fracture surface of 1200°C austenitized, 500°C tempered 5Mo-.3C steel. Quasi-cleavage visible. $\Delta K=31 \text{ ksi-in.}^{1/2}$.
- Fig. 22. Fast fracture appearance of 1200°C austenitized, 500°C tempered 5Mo-.3C steel showing quasi-cleavage appearance.
- Fig. 23. Microstructure of 1200°C austenitized, 600°C tempered 5Mo-.3C steel.
- Fig. 24. Composite SEM fractograph showing fatigue fracture surface of 1200°C austenitized, 600°C tempered 5Mo-.3C steel. Intergranular mode of failure evident in fatigue area. $\Delta K=21 \text{ ksi-in.}^{1/2}$.
- Fig. 25. Fast fracture appearance of 1200°C, austenitized 600°C tempered 5Mo-.3C steel. (a) Intergranular modes shown (b) Quasi-cleavage shown.
- Fig. 26. Hardness vs tempering temperature for 5Mo-.3C steel austenitized at 870°C.
- Fig. 27. Yield strength as a function of tempering temperature for 5Mo-.3C steel austenitized at 870°C.

Fig. 28. Fracture toughness as a function of tempering temperature for 5Mo-.3C steel austenitized at 870°C.

Fig. 29. Fatigue crack propagation rates vs stress intensity range for 870°C austenitized, as-quenched 5Mo-.3C steel. Data from specimen 18-2(6).

Fig. 30. Fatigue crack propagation rates vs stress intensity range for 870°C austenitized, 225°C tempered 5Mo-.3C steel. Data from specimen 18-3(6).

Fig. 31. Fatigue crack propagation rates vs stress intensity range for 870°C austenitized, 300°C tempered 5Mo-.3C steel. Data from specimens 15-5(6) and 18-6(6).

Fig. 32. Fatigue crack propagation rates vs stress intensity range for 870°C austenitized, 500°C tempered 5Mo-.3C steel. Data from specimens 18-7(6) and 15-8(6).

Fig. 33. Fatigue crack propagation rates vs stress intensity range for 870°C austenitized, 600°C tempered 5Mo-.3C steel. Data from specimens 15-9(6) and 18-0(6).

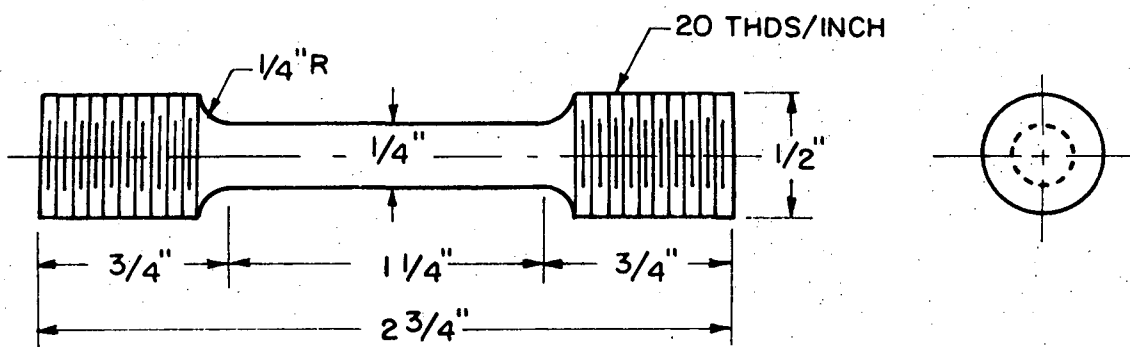
Fig. 34. Microstructure of 870°C austenitized, as-quenched 5Mo-.3C steel. (a) High fraction of carbides shown (b) Light areas, possibly ferrite shown.

Fig. 35. Fatigue fracture region bordering fast fracture region for 870°C austenitized, as-quenched, 5Mo-.3C steel. (a) Overall appearance. (b) Composite enlargement of region.

Fig. 36. Fast fracture appearance of 870°C austenitized, as-quenched, 5Mo-.3C steel. (a) Overall appearance. (b) Enlargement showing quasi-cleavage appearance.

- Fig. 37. Microstructure of as-quenched 5Mo-.3C steel after austenitization at 980°C. Undissolved carbides present.
- Fig. 38. Fatigue crack growth rates vs stress intensity range for 980°C austenitized, as-quenched 5Mo-.3C steel. Data from specimen 1800-1.
- Fig. 39. Fatigue fracture appearance in as-quenched 5Mo-.3C steel that was austenitized at 980°C.
- Fig. 40. Fast fracture appearance for 980°C austenitized, as-quenched 5Mo-.3C steel. Quasi-cleavage is primary feature.
- Fig. 41. Microstructure for 1Mo-.24C steel. Material was austenitized at 870°C, ice-brine quenched.
- Fig. 42. Microstructure for 870°C austenitized, as-quenched 1Mo-.34C steel.
- Fig. 43. Fatigue crack growth rates vs stress intensity range for 870°C austenitized, as-quenched 1Mo-.3C steel. Data from specimen 321.
- Fig. 44. Fatigue crack growth rates vs stress intensity range for 870°C austenitized, as-quenched 1Mo-.34C steel. Data from specimen 324.
- Fig. 45. Fatigue fracture appearance for specimen 321.
- Fig. 46. Fatigue fracture appearance for specimen 324.
- Fig. 47. Fast fracture appearance for specimen 321. Dimpled rupture characteristic of surface.
- Fig. 48. Fast fracture appearance for specimen 324.

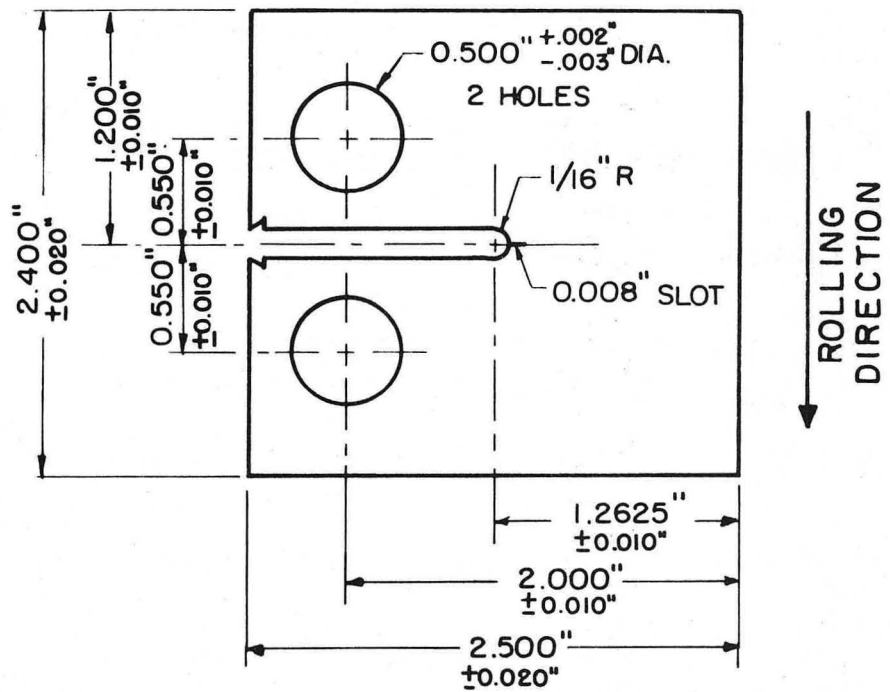
- Fig. 49. Comparison of fatigue crack growth and fracture toughness properties for the different tempering treatments given the 5Mo-0.3C steel after austenitization at 1200°C. Number of cycles to grow one inch at a constant stress intensity range are used to measure crack growth characteristics.
- Fig. 50. Comparisons of fatigue crack propagation rate vs stress intensity range of 1200°C austenitized as-quenched 5Mo-.3C steel with rates reported for other high-strength steels (A-Ref. 10, B-Ref. 12, C-Ref. 35 and D-Ref. 9).
- Fig. 51. Comparison of fatigue crack growth rates vs stress intensity range for as-quenched 5Mo-.3C steel.
- Fig. 52. Comparison of number of cycles to failure for different loads for 5Mo-.3C steel. Life determined by integrating Paris relation.



(b) ROUND TENSILE SPECIMEN

XBL 735-6189A

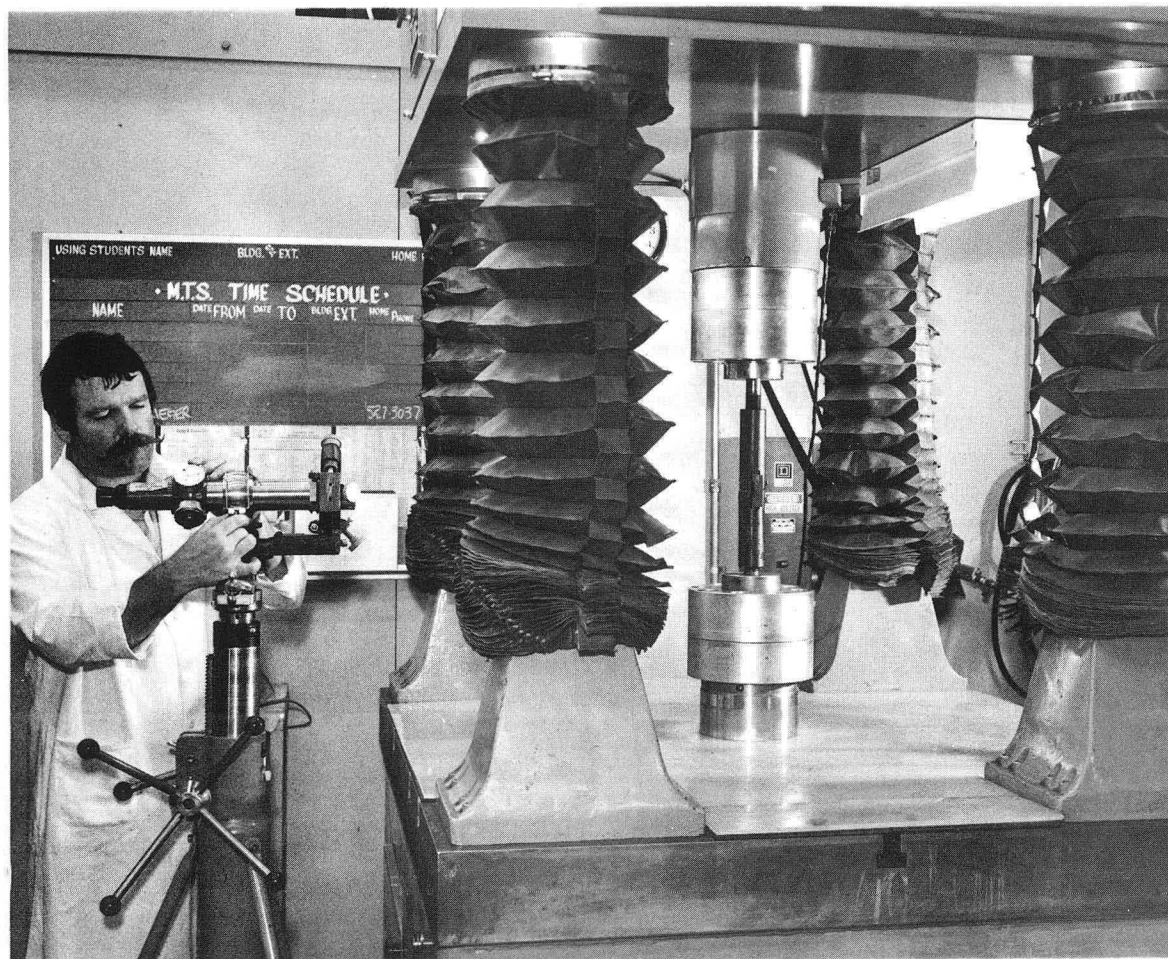
Fig. 1.



FRACTURE TOUGHNESS SPECIMEN

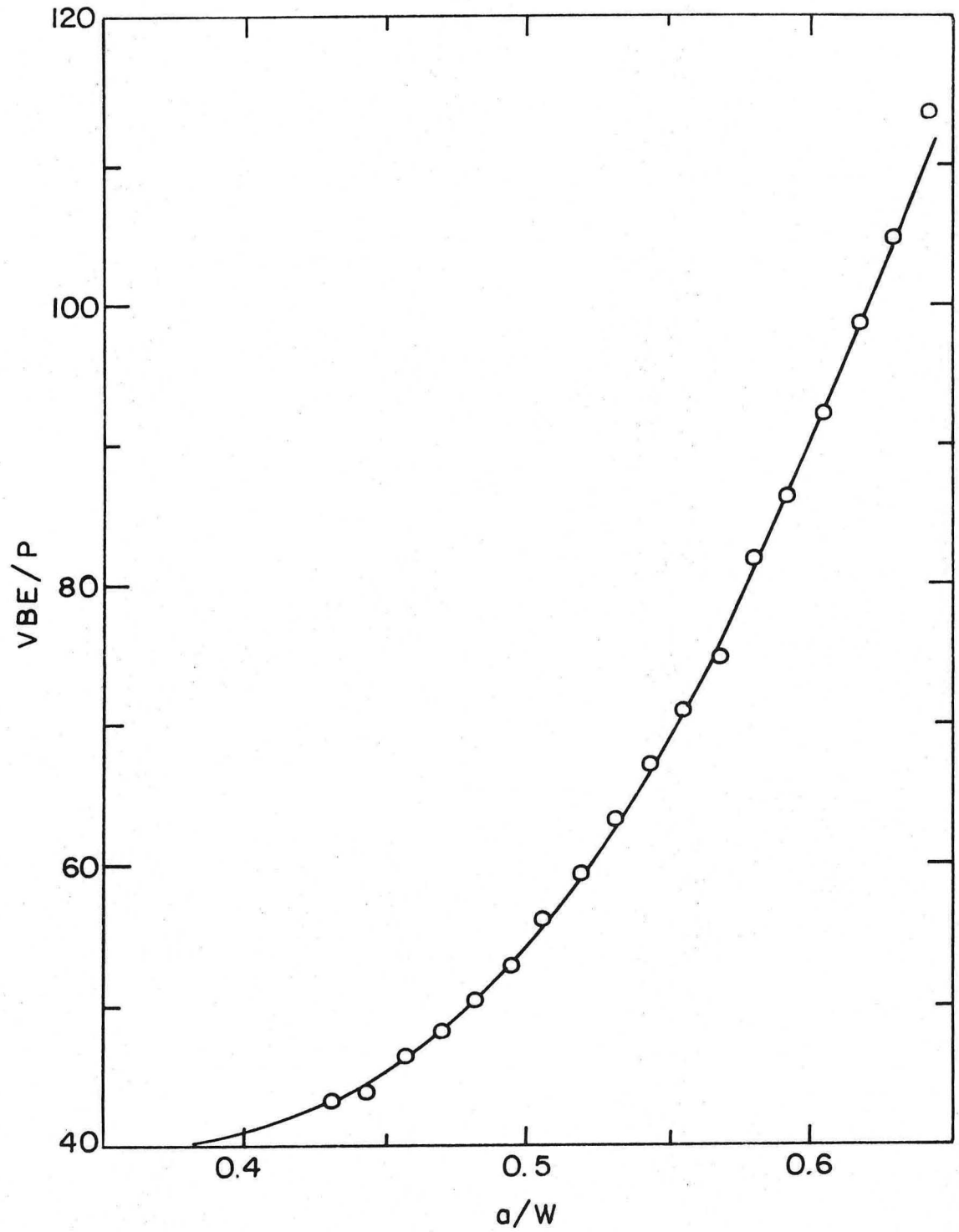
XBL 737- 6465

Fig. 2.



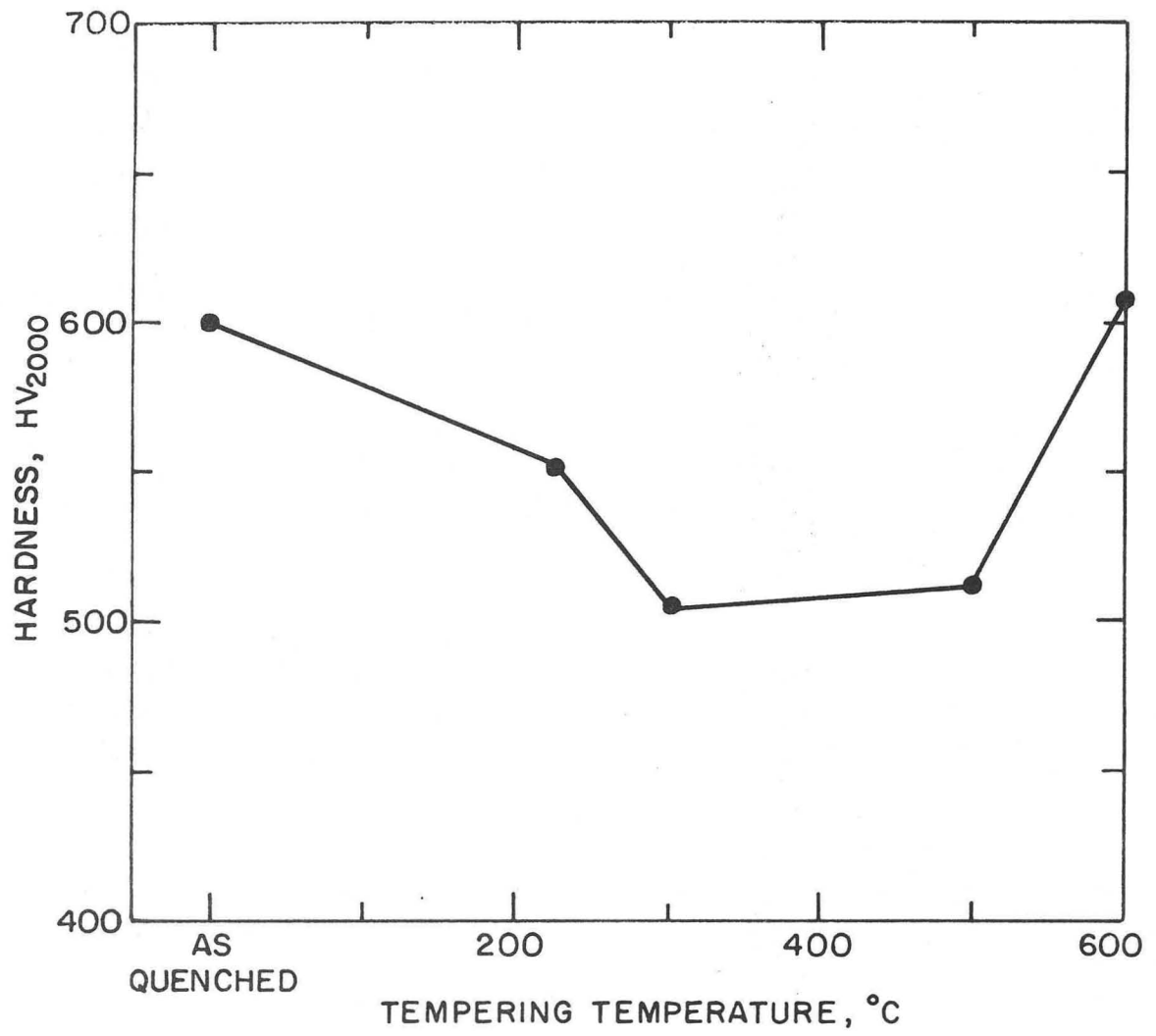
XBB 7311-6572

Fig. 3.



XBL737 - 6467

Fig. 4.



XBL 7311-660I

Fig. 5.

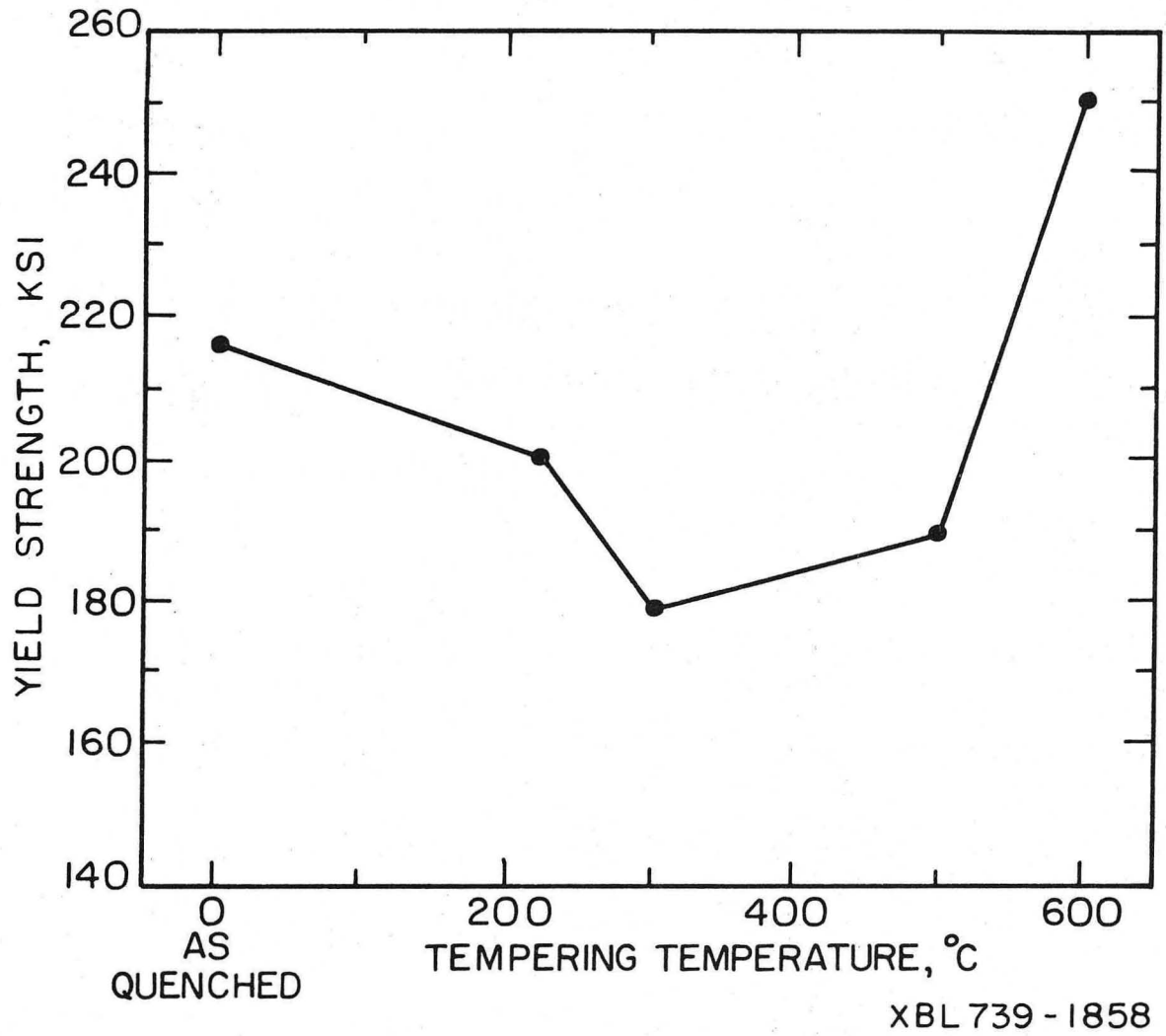
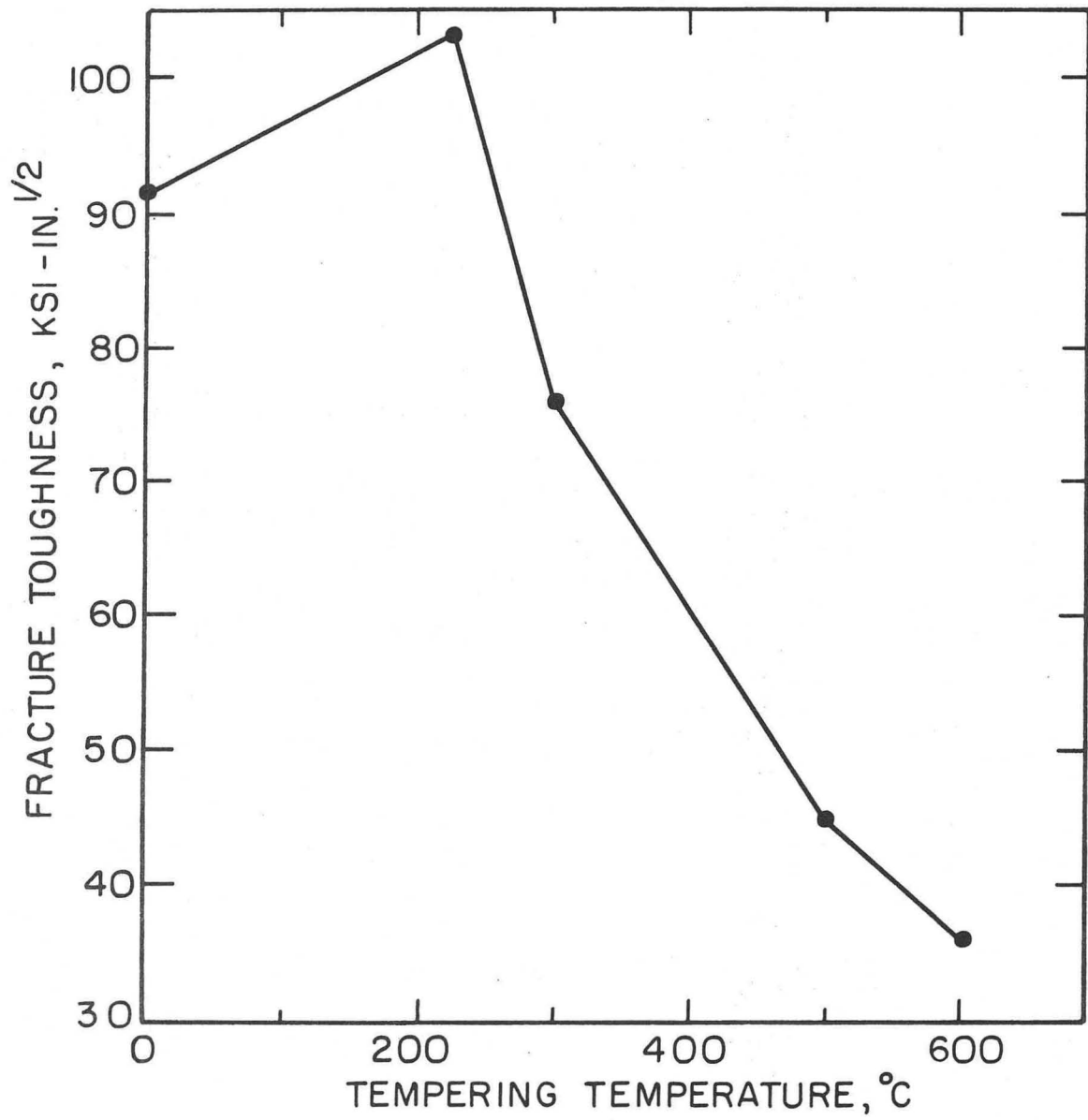


Fig. 6.



XBL 739-1859

Fig. 7.

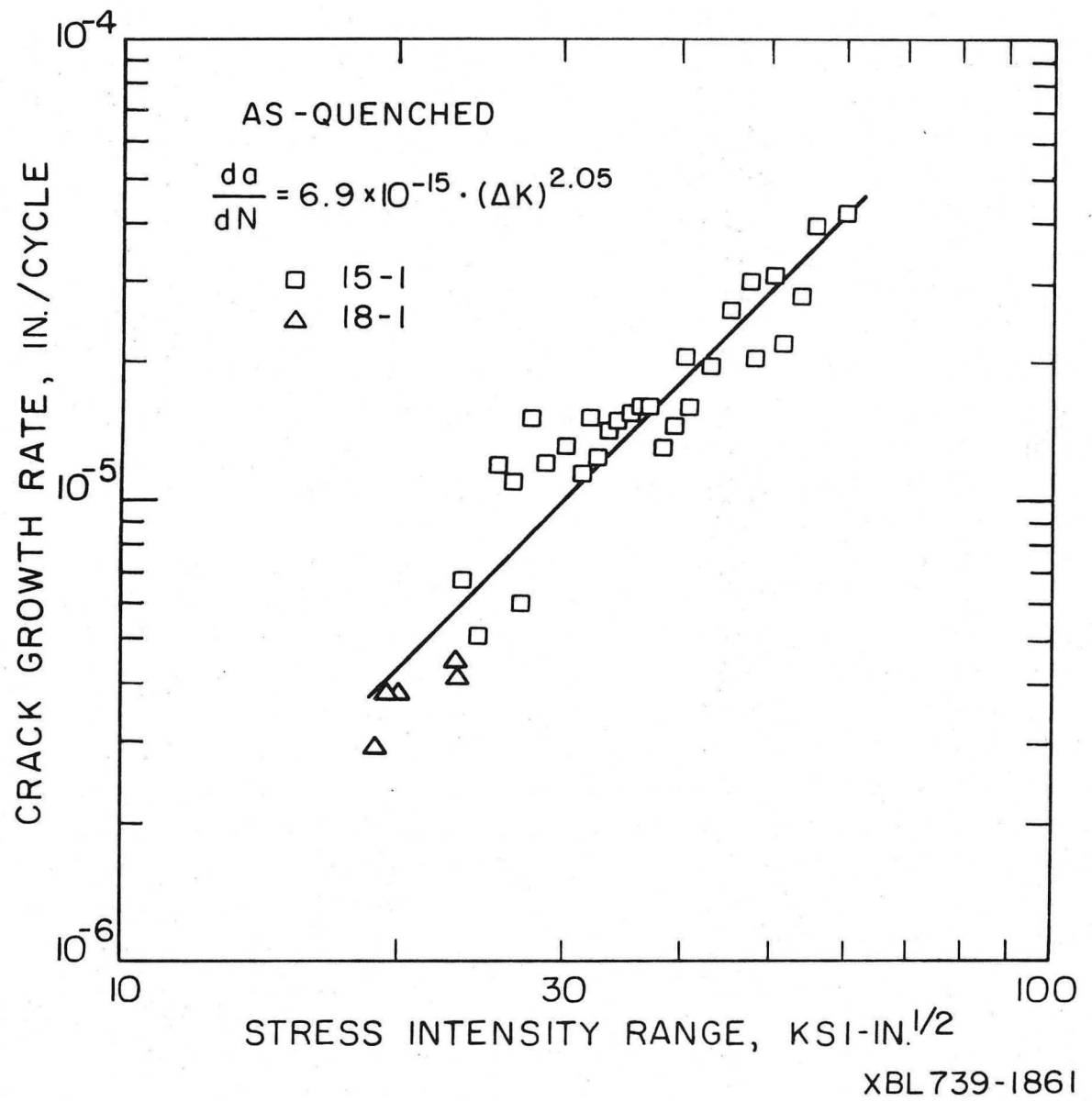
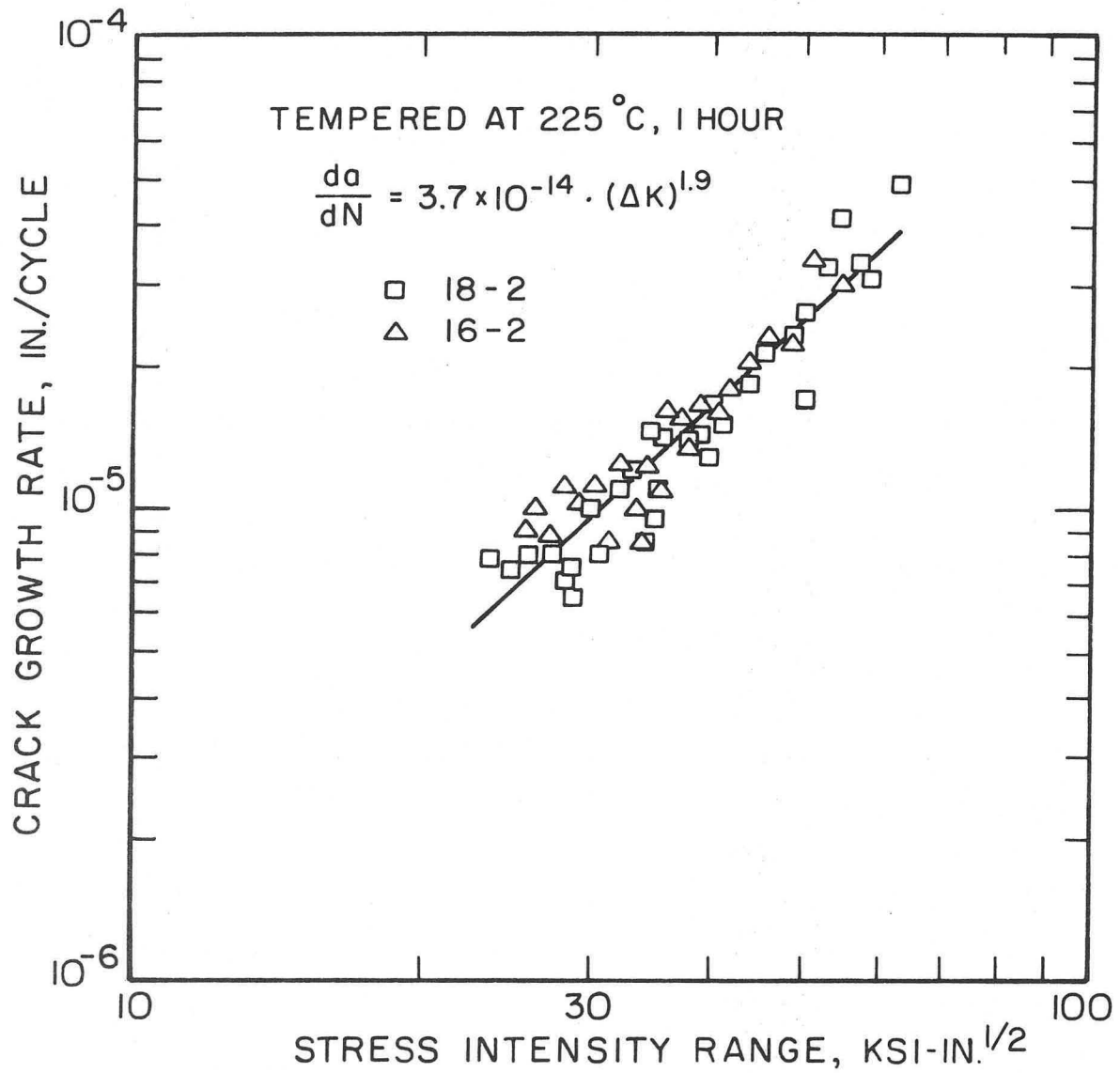
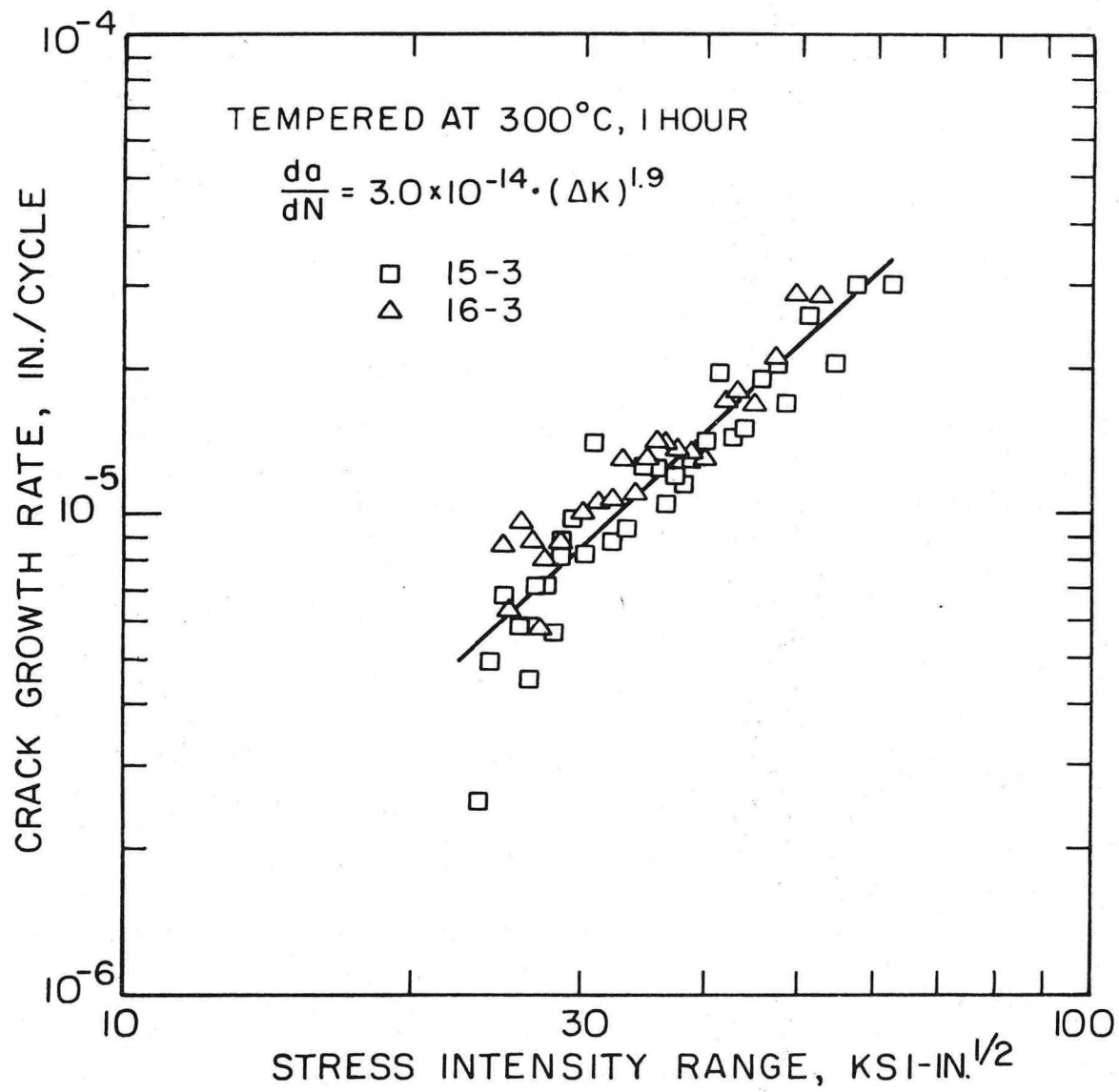


Fig. 8.



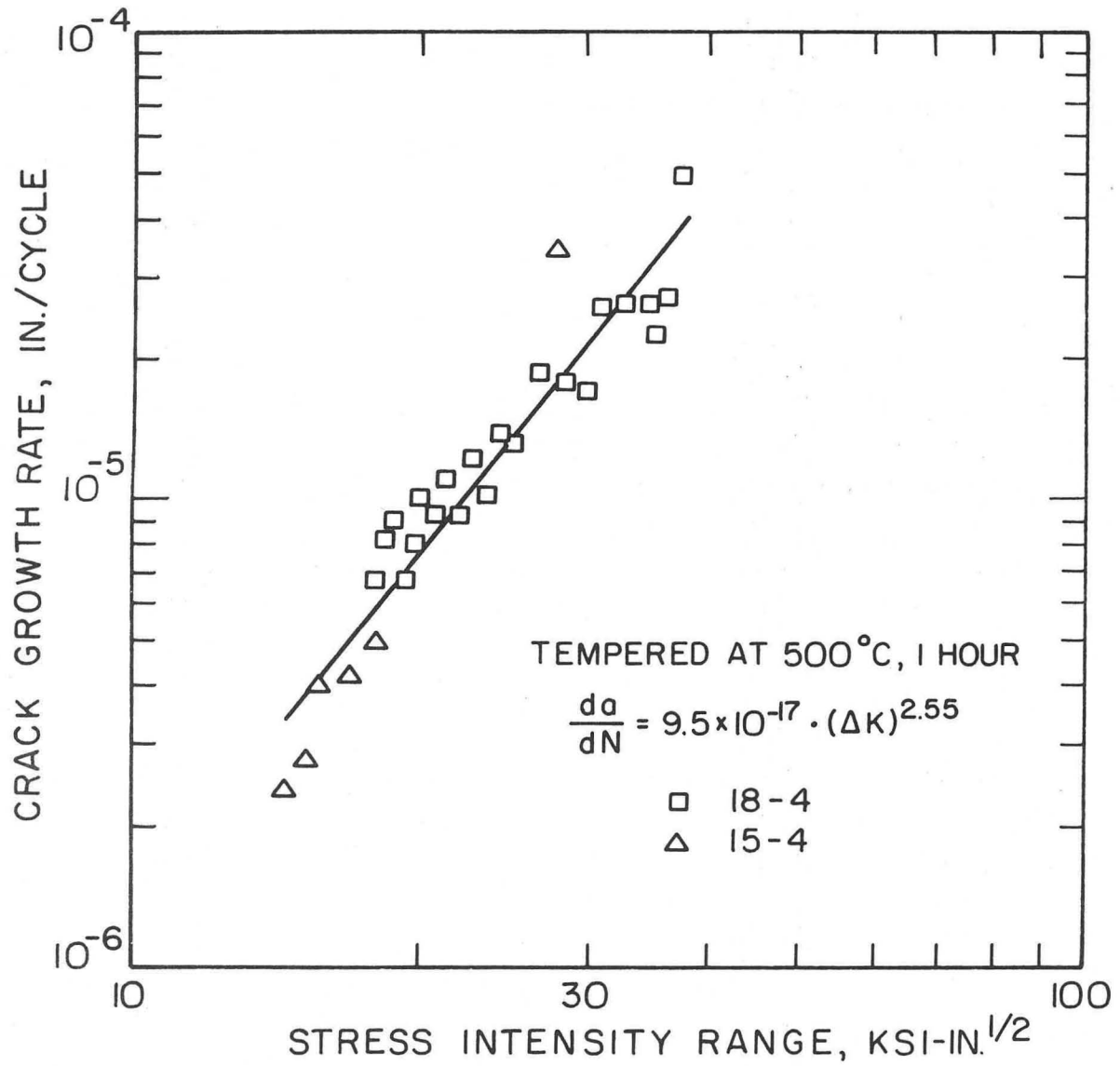
XBL 739-1862

Fig. 9.



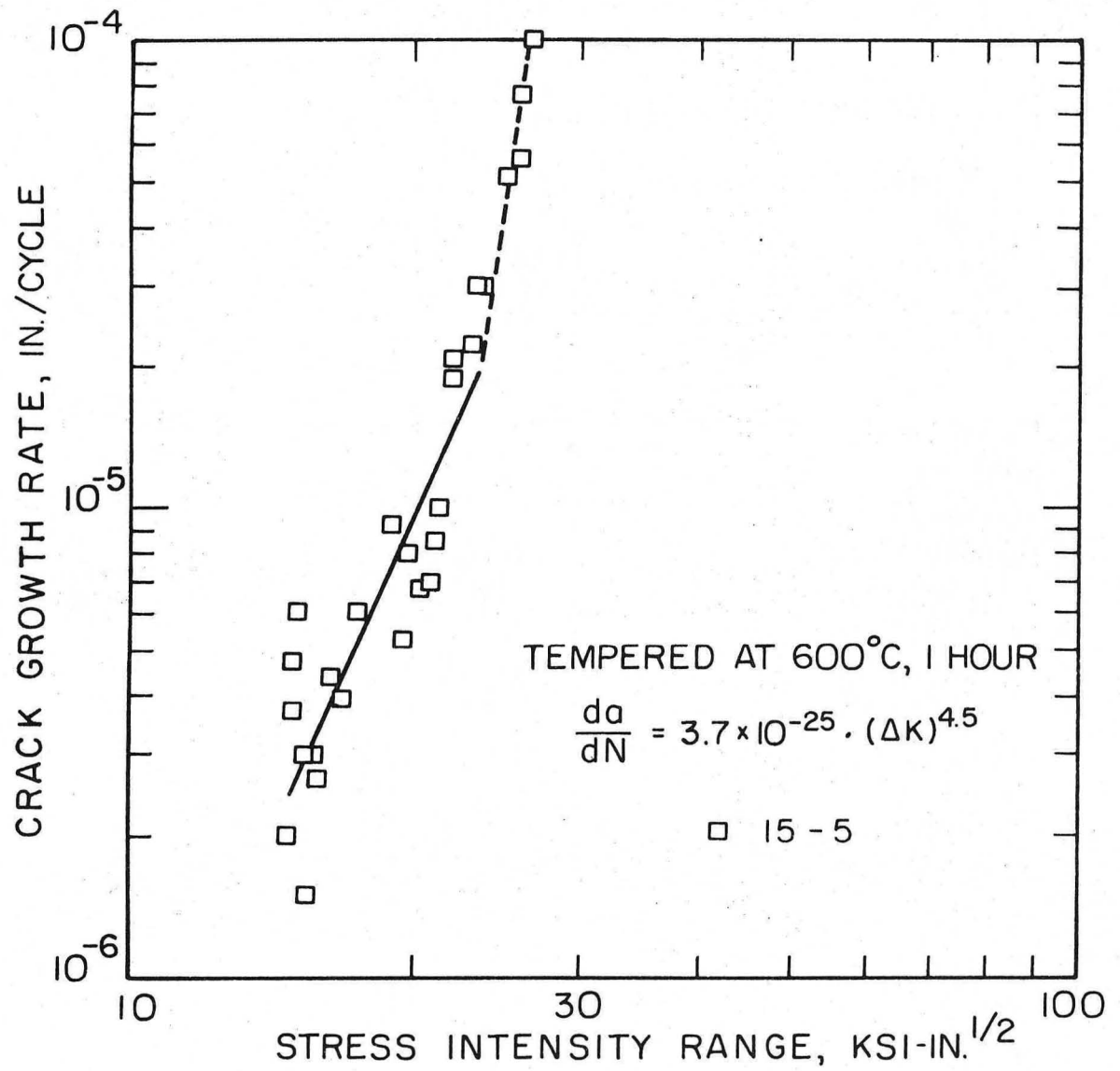
XBL 739-1863

Fig. 10.



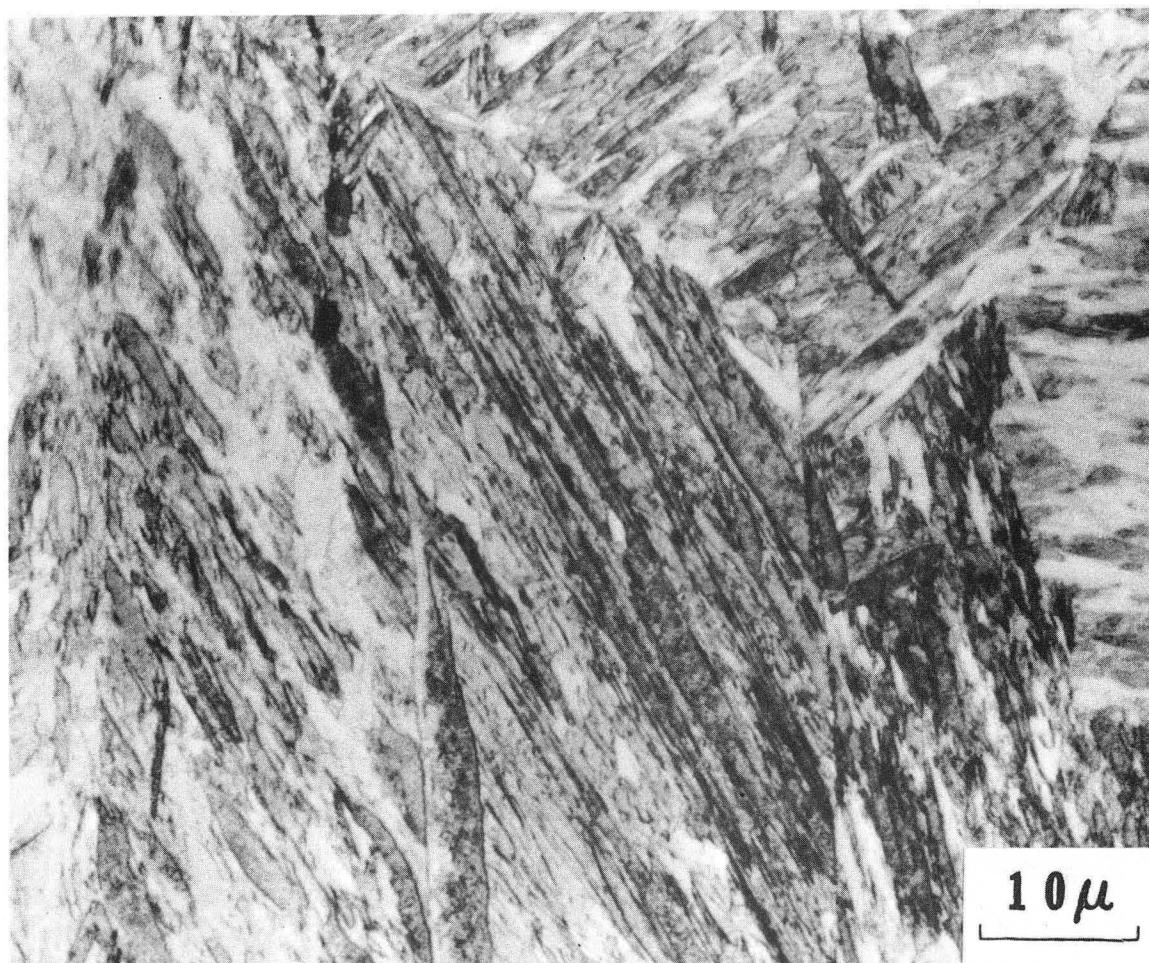
XBL 739-1864

Fig. 11.



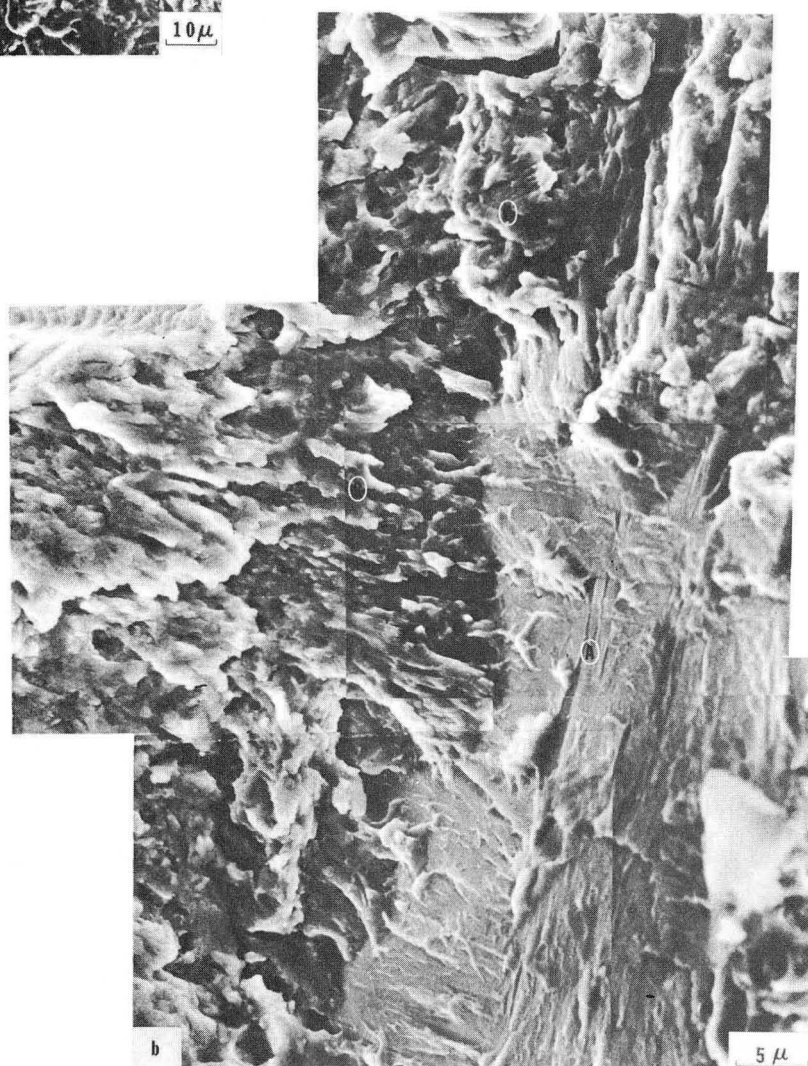
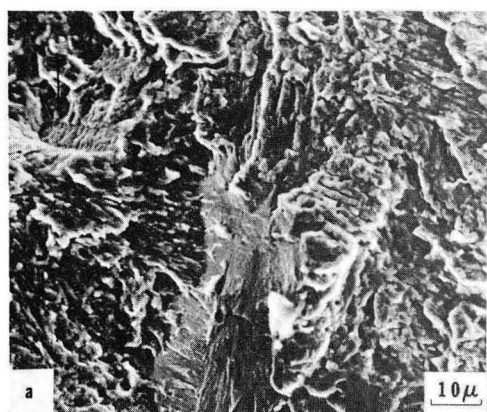
XBL 739-1865

Fig. 12.



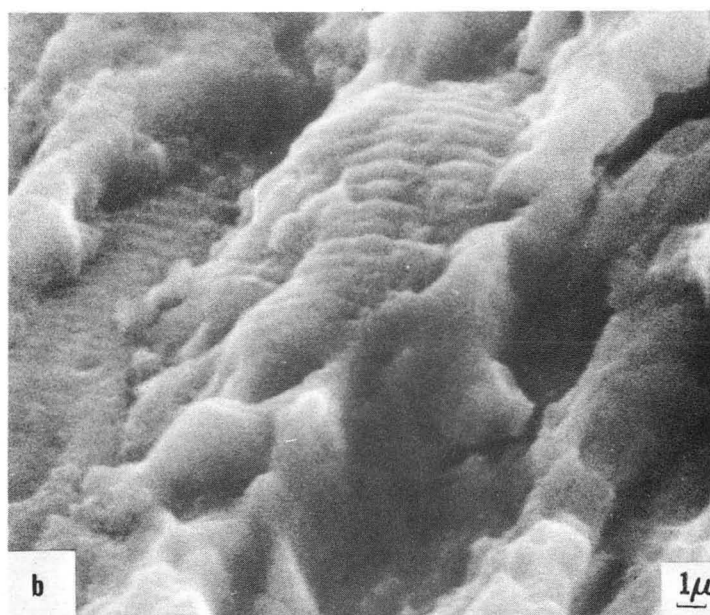
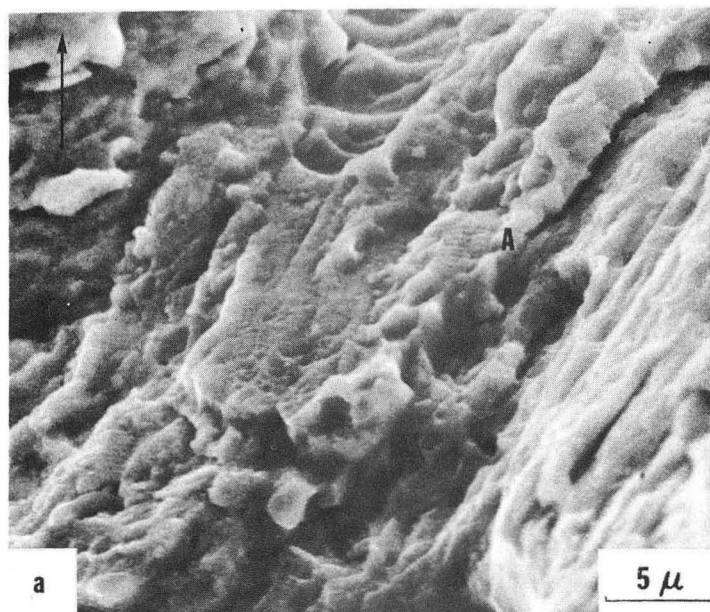
XBB 739-5584

Fig. 13.



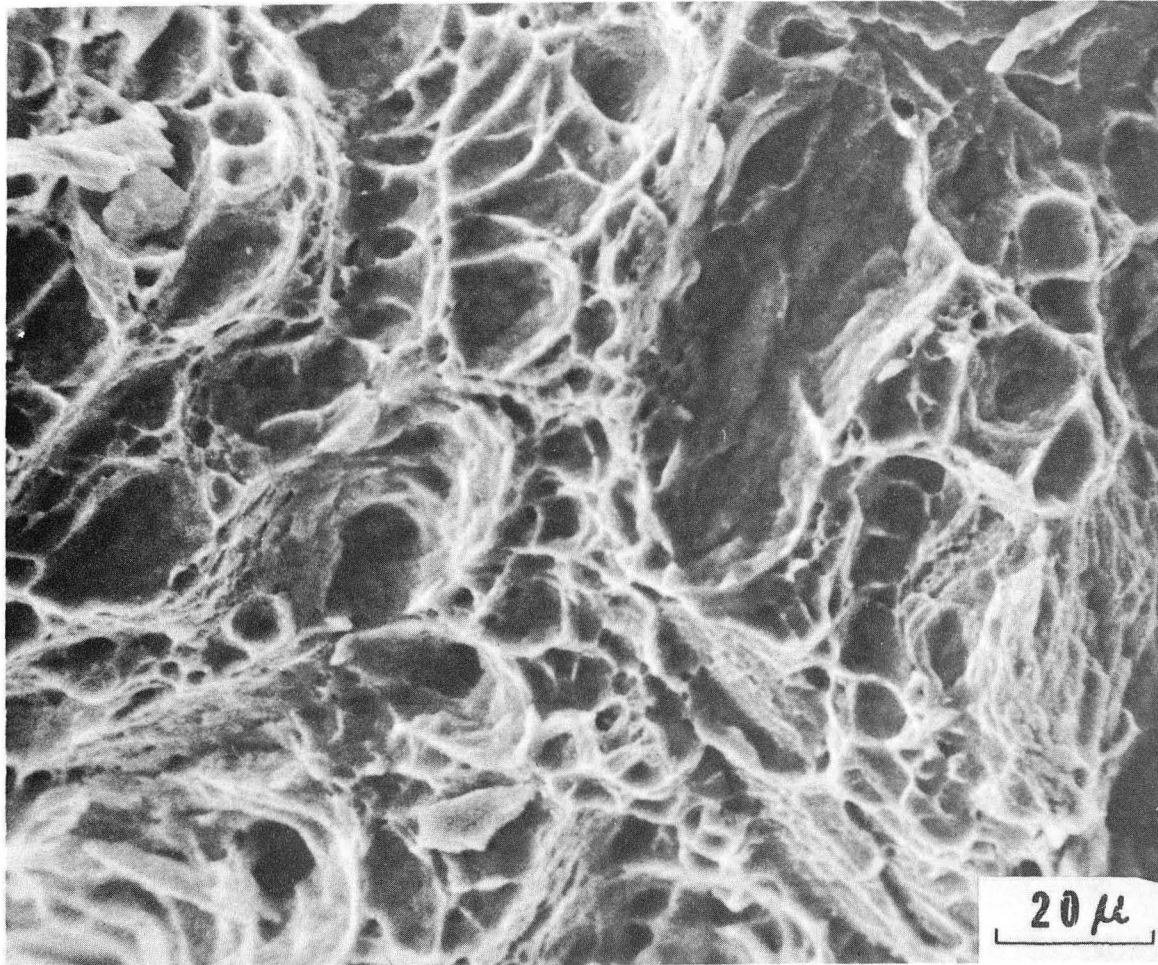
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Fig. 14.



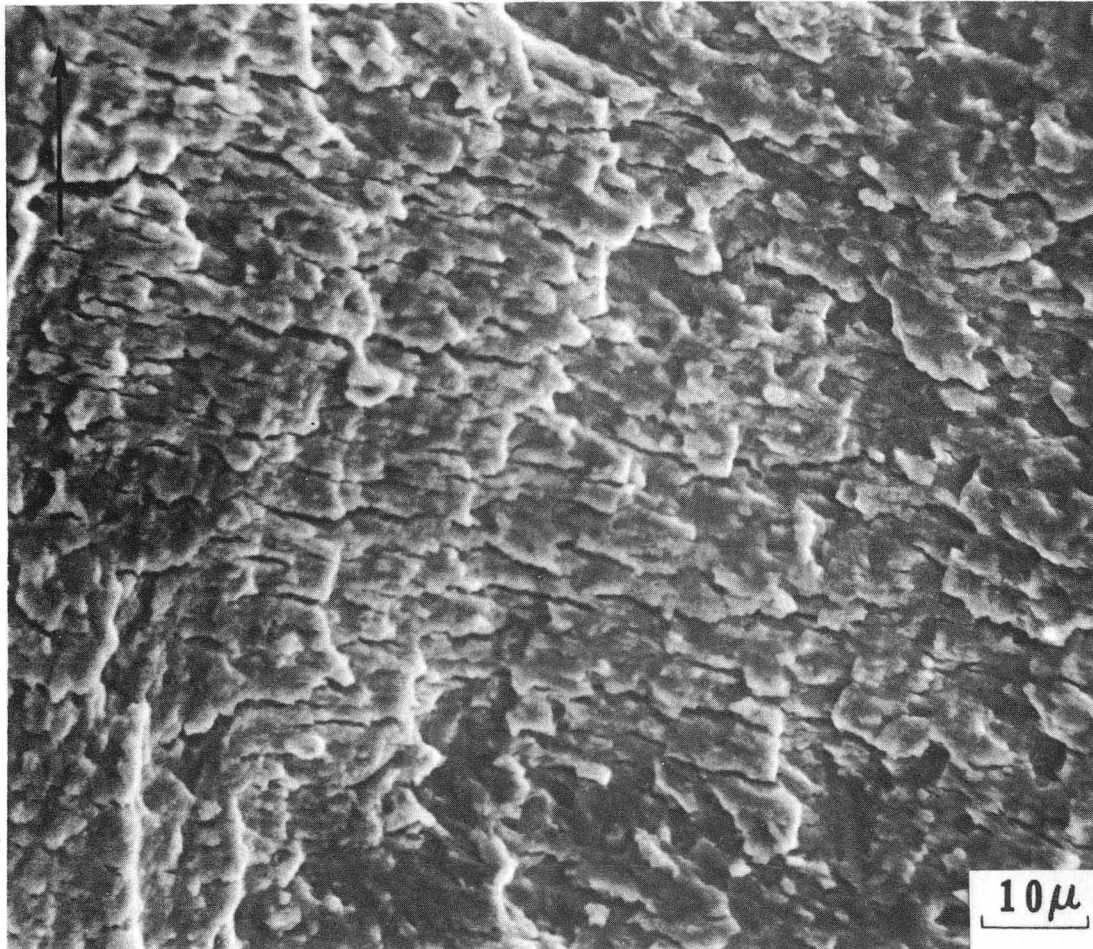
XBB 739-5581

Fig. 15.



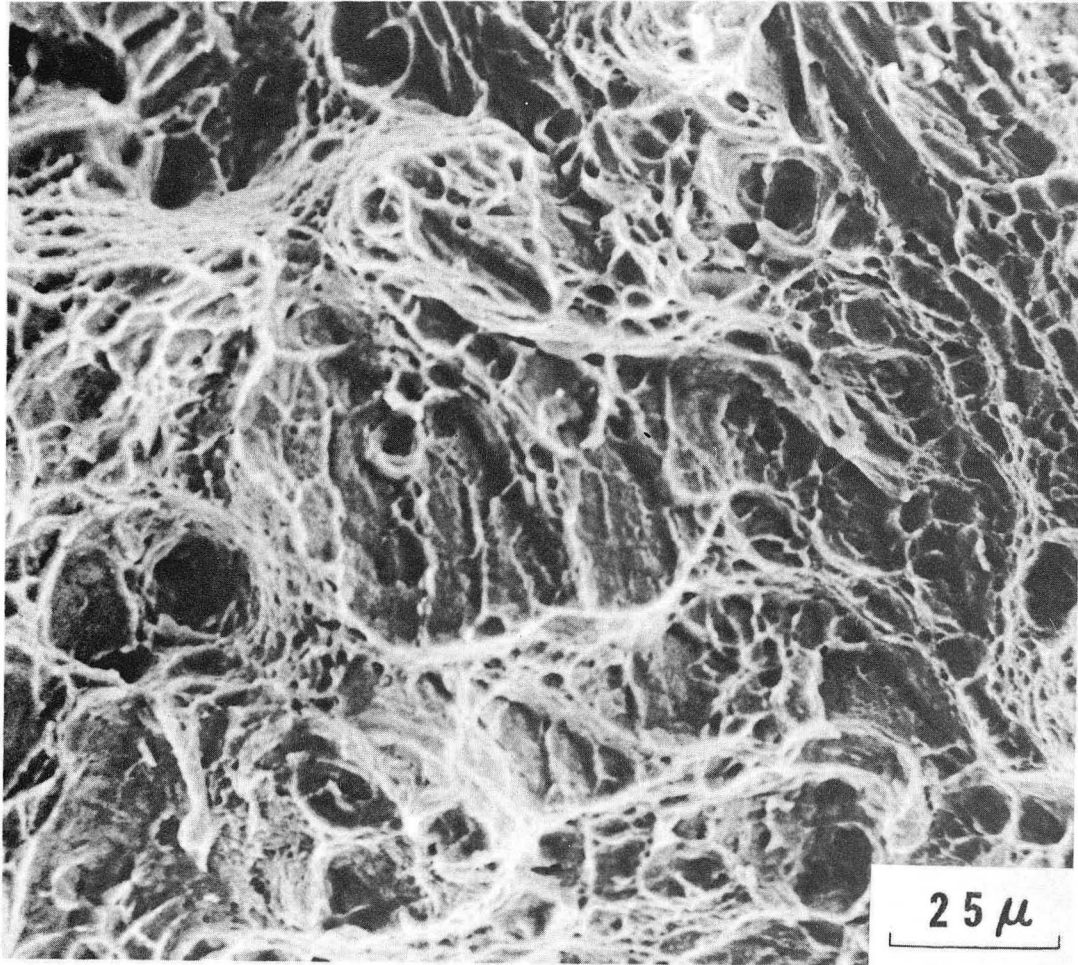
XBB 7311-6689

Fig. 16.



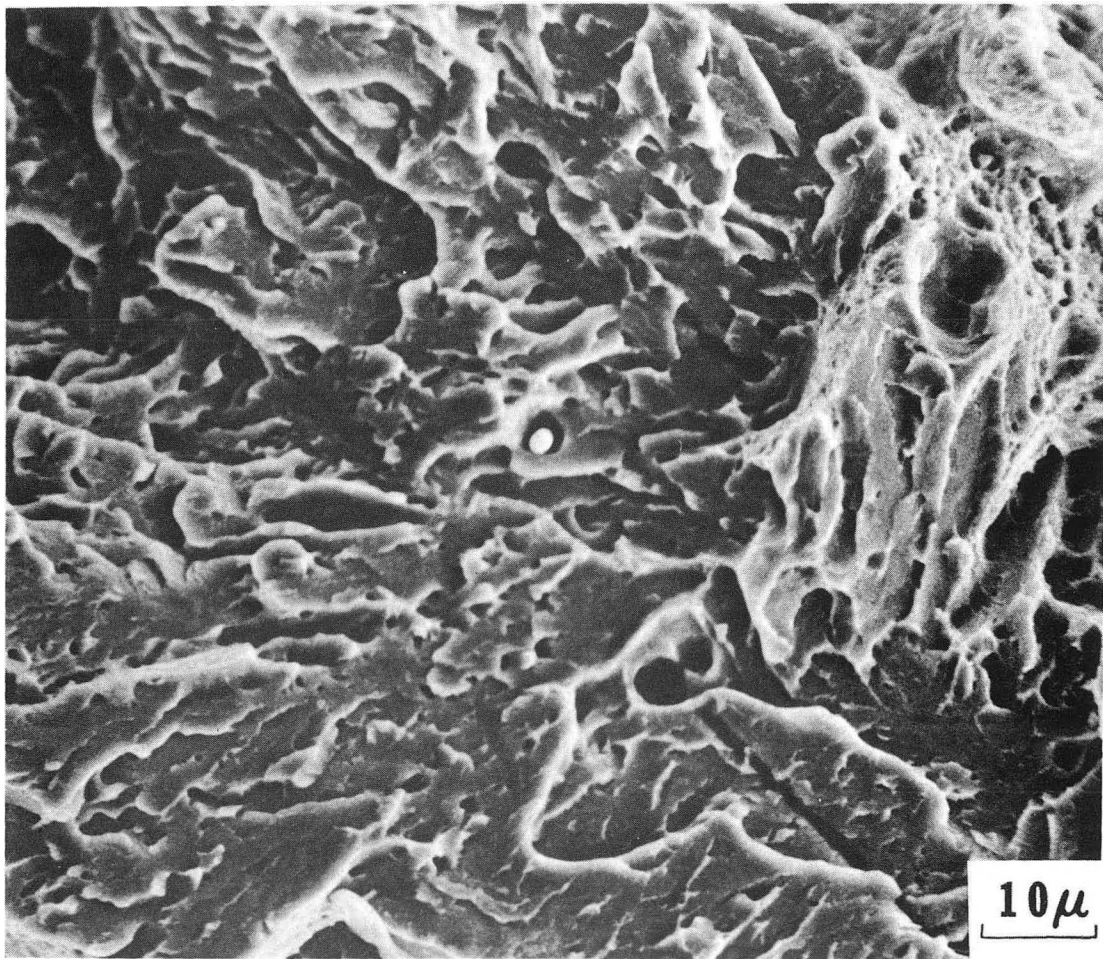
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Fig. 17.



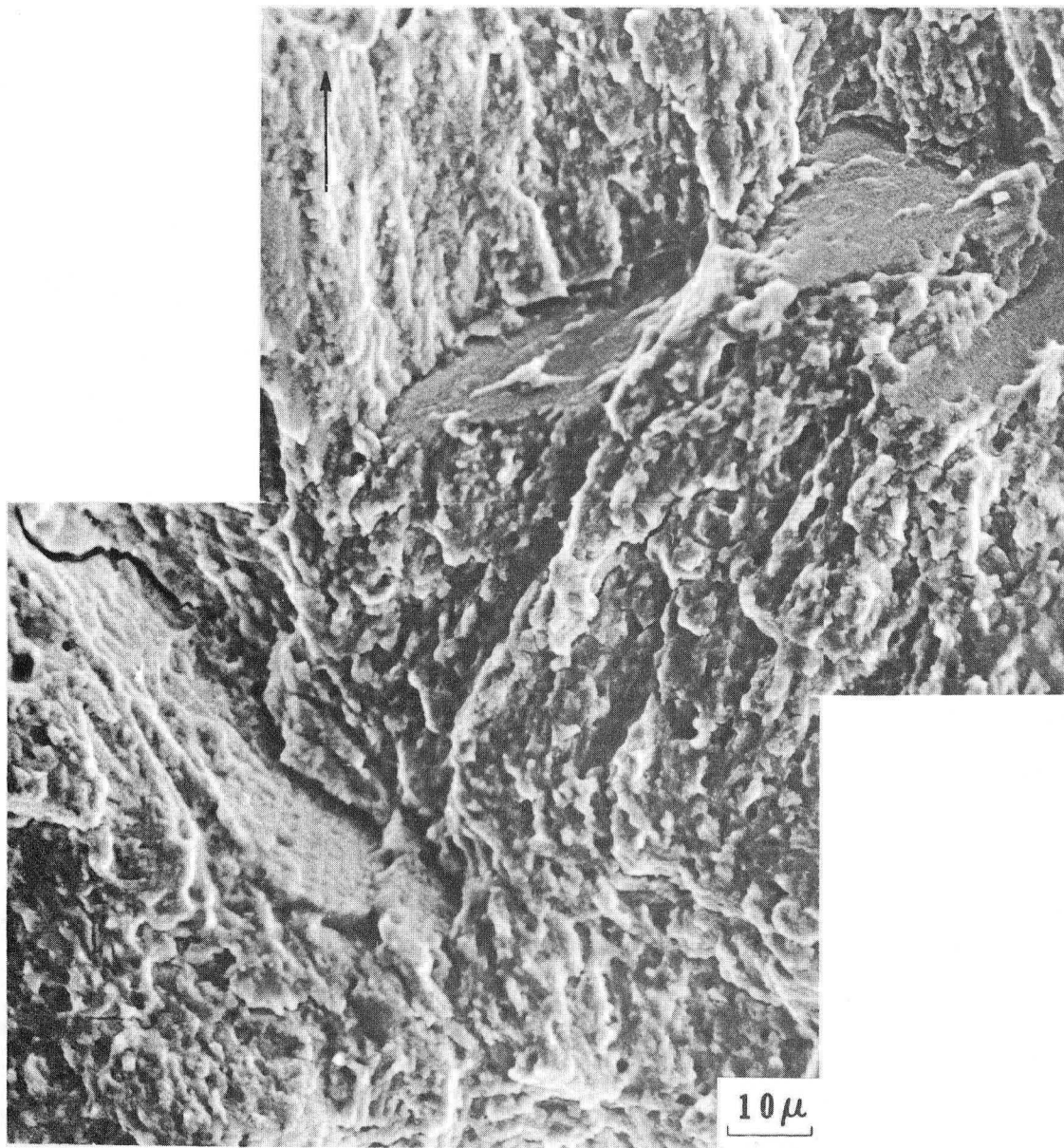
XBB 739-5587

Fig. 18.



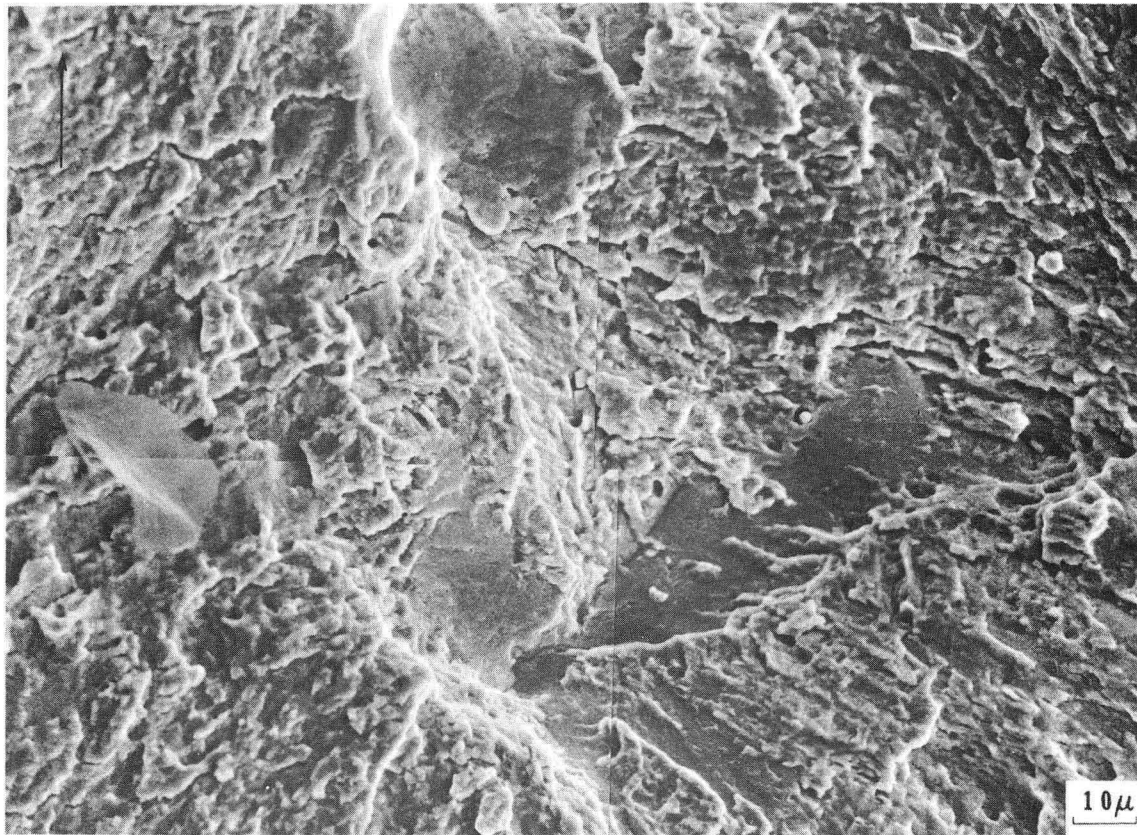
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Fig. 19.



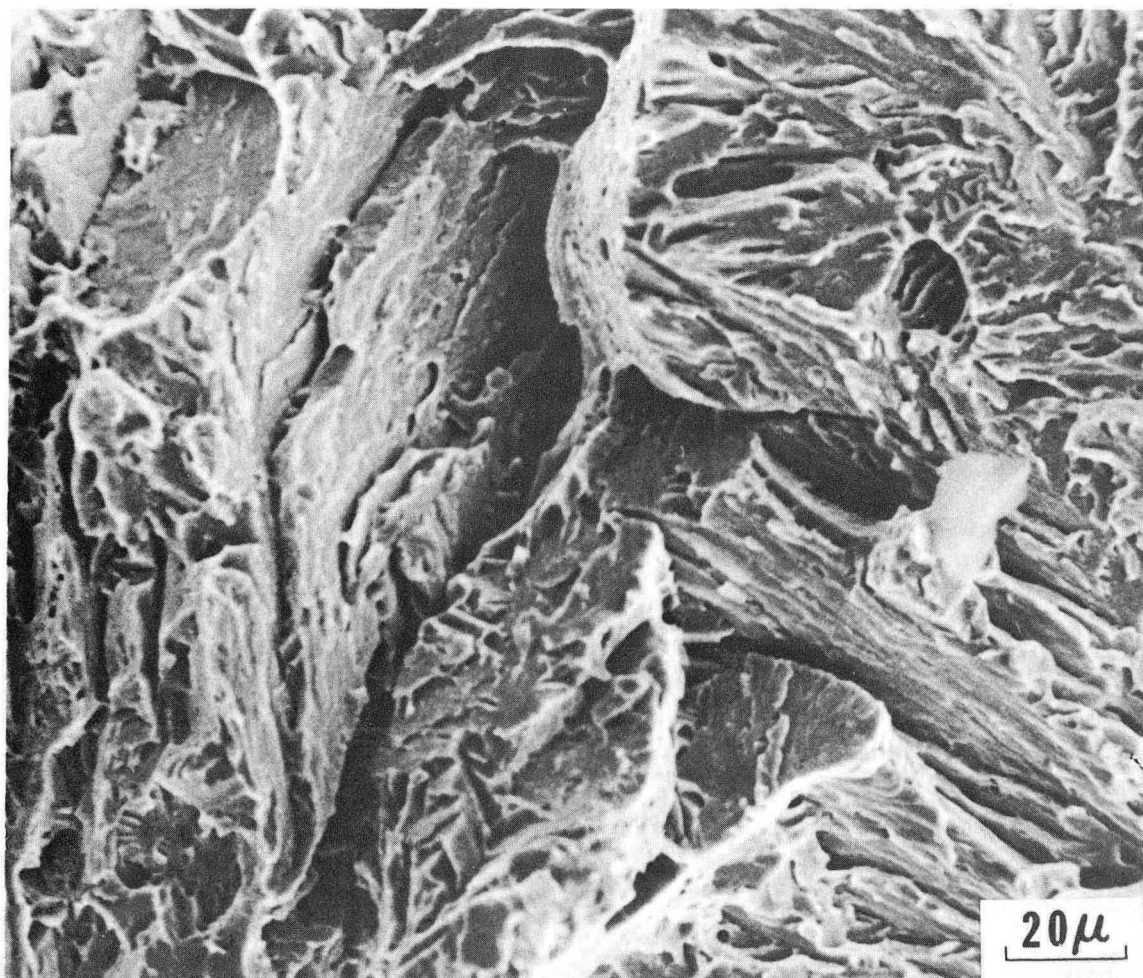
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Fig. 20.



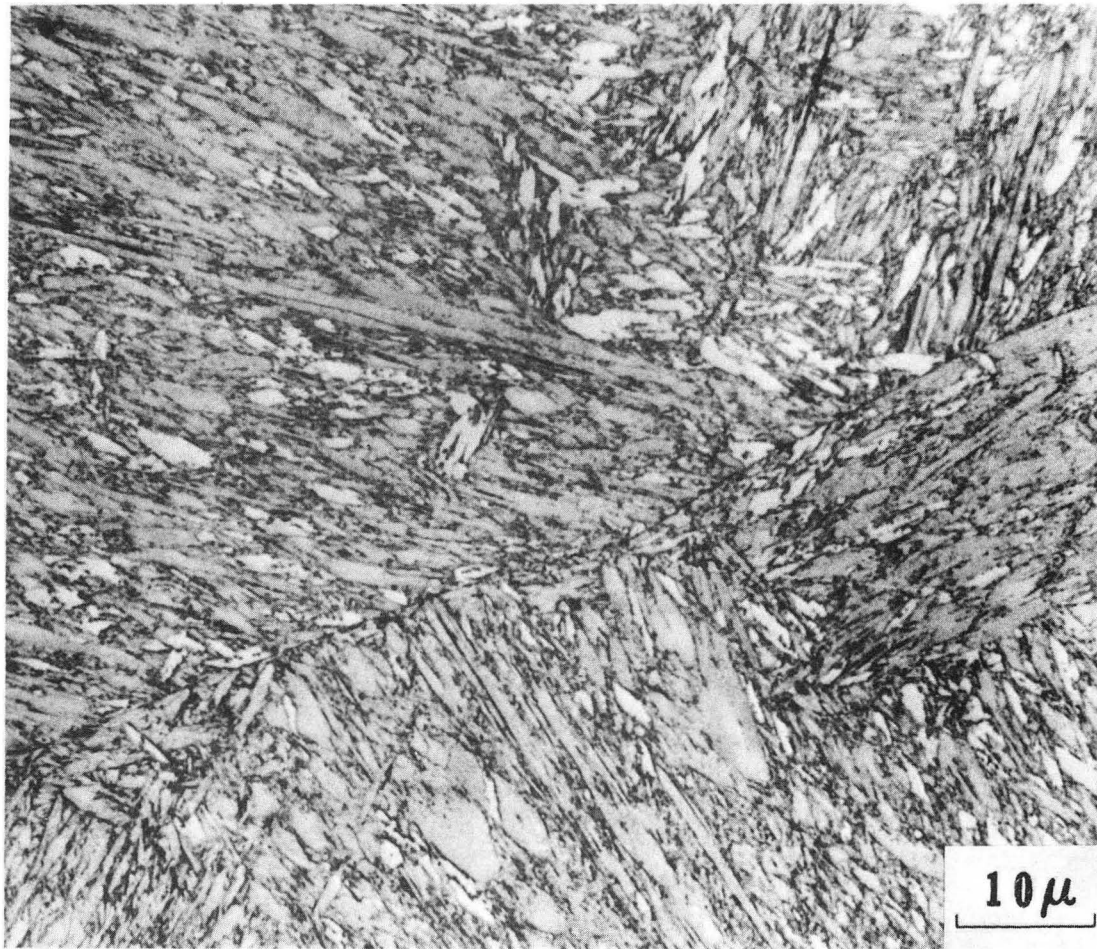
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Fig. 21.



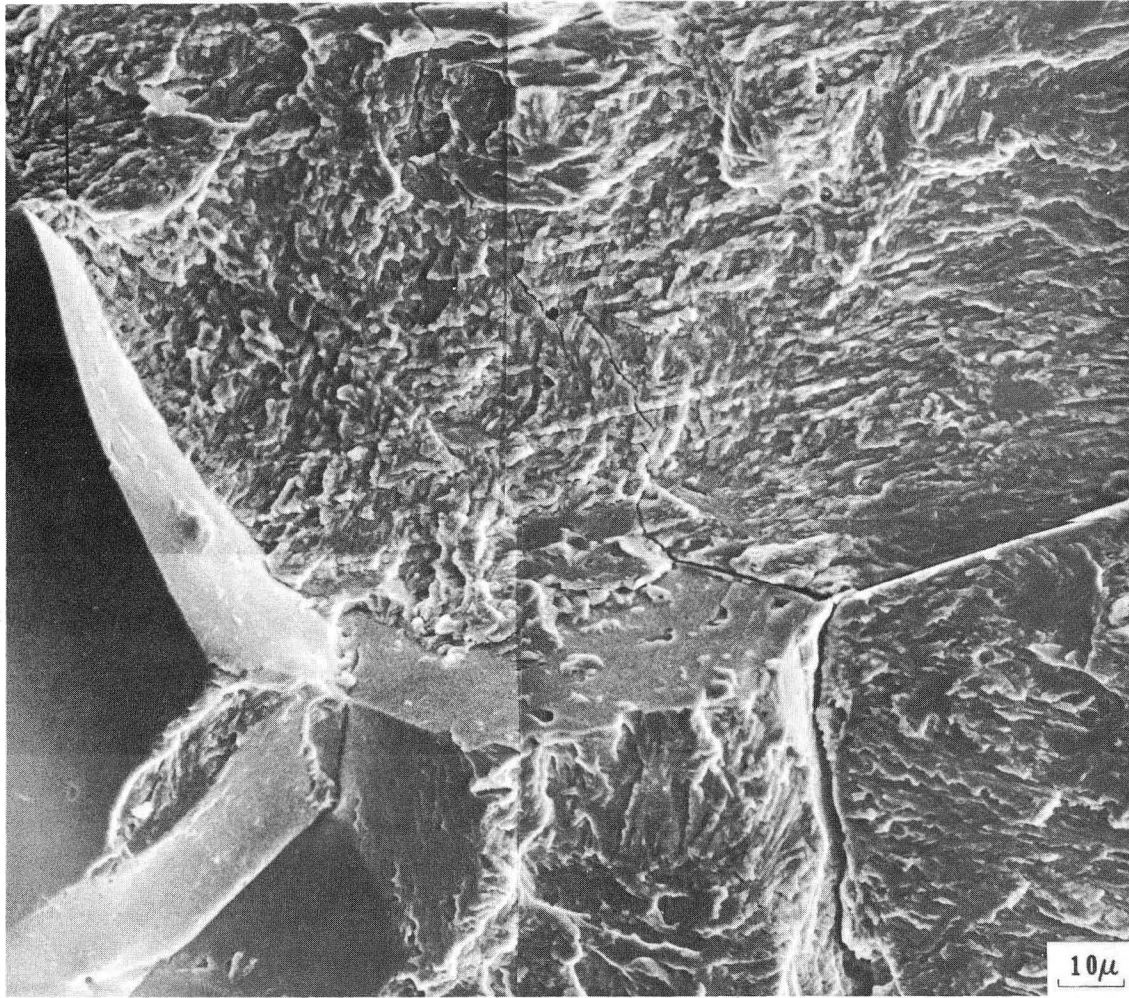
XBB 7311-6687

Fig. 22.



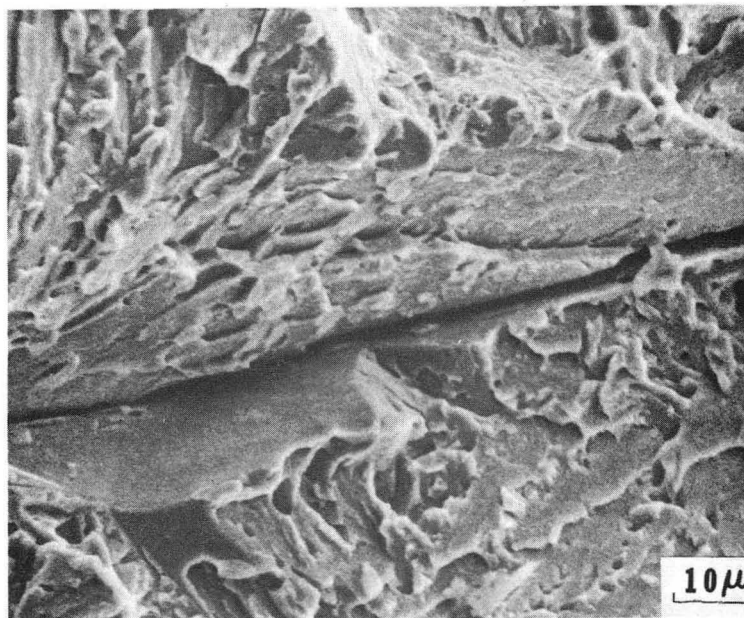
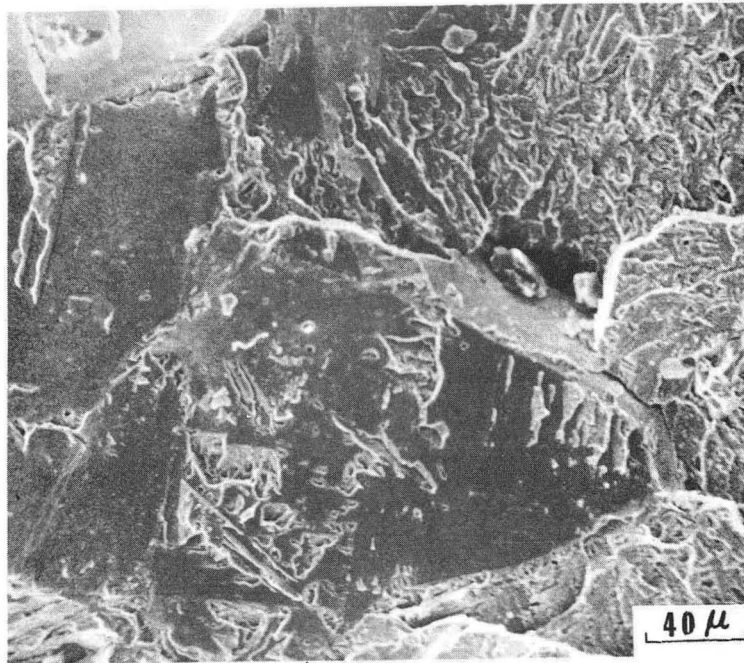
XBB 739-5585

Fig. 23.



XBB 739-5583

Fig. 24.



XBB 7311-6682

Fig. 25.

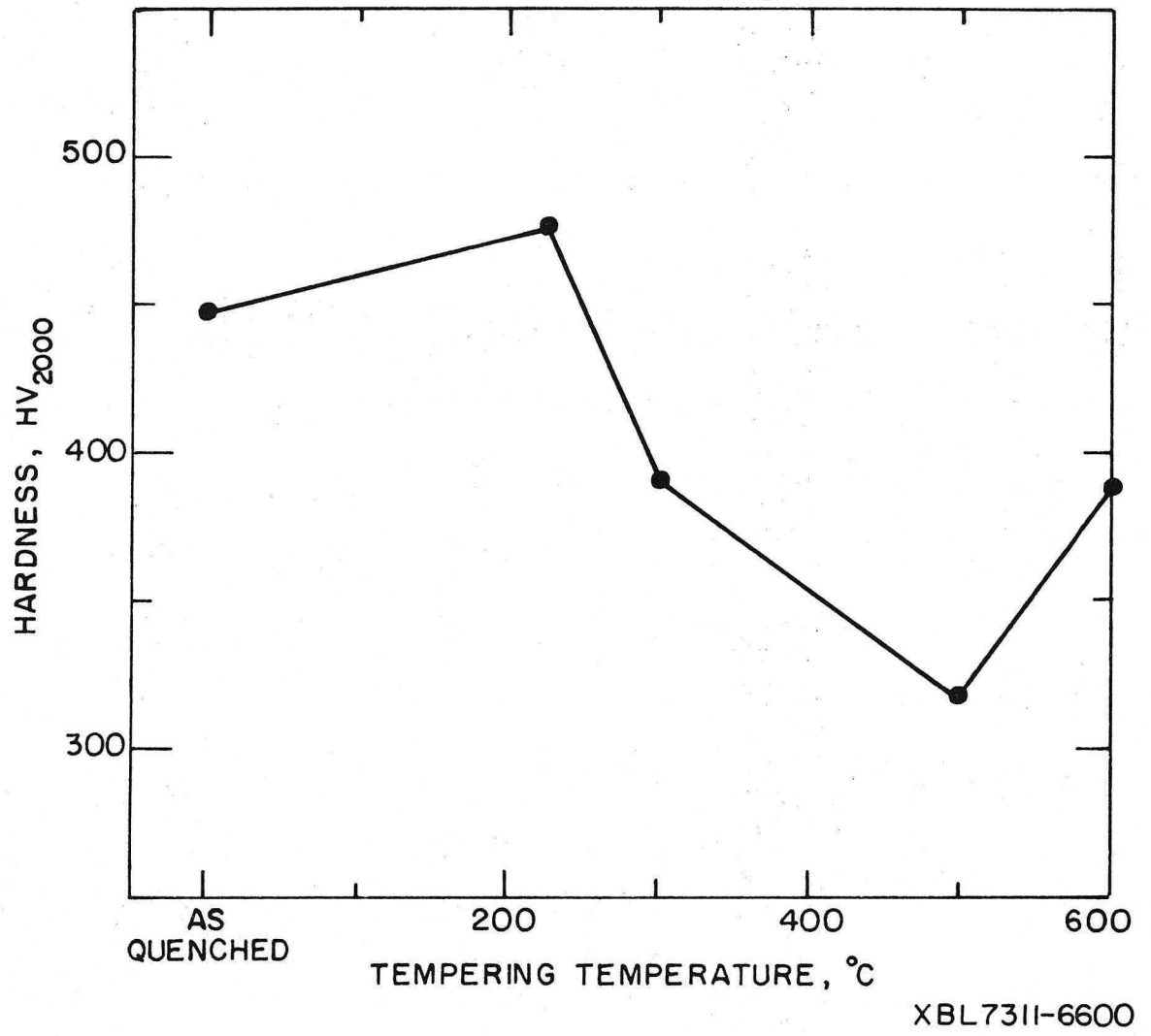
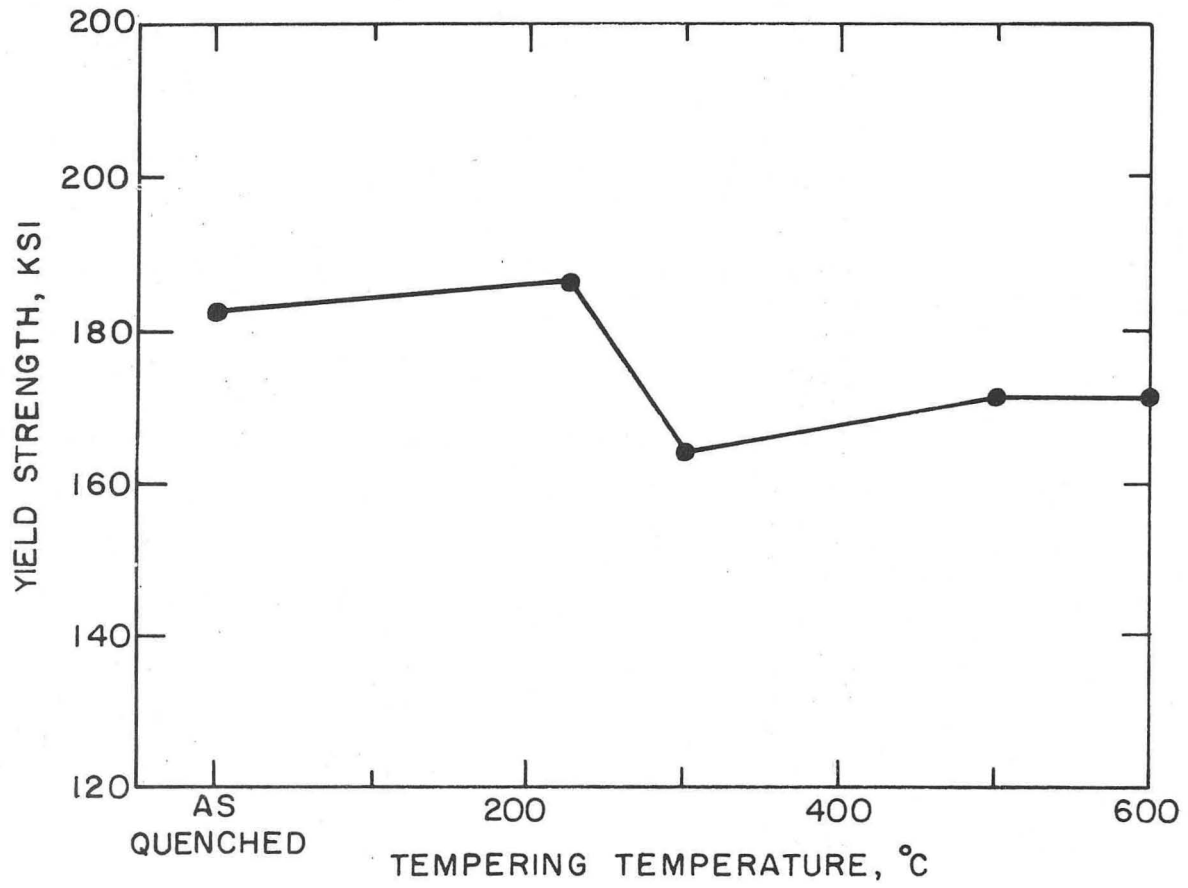
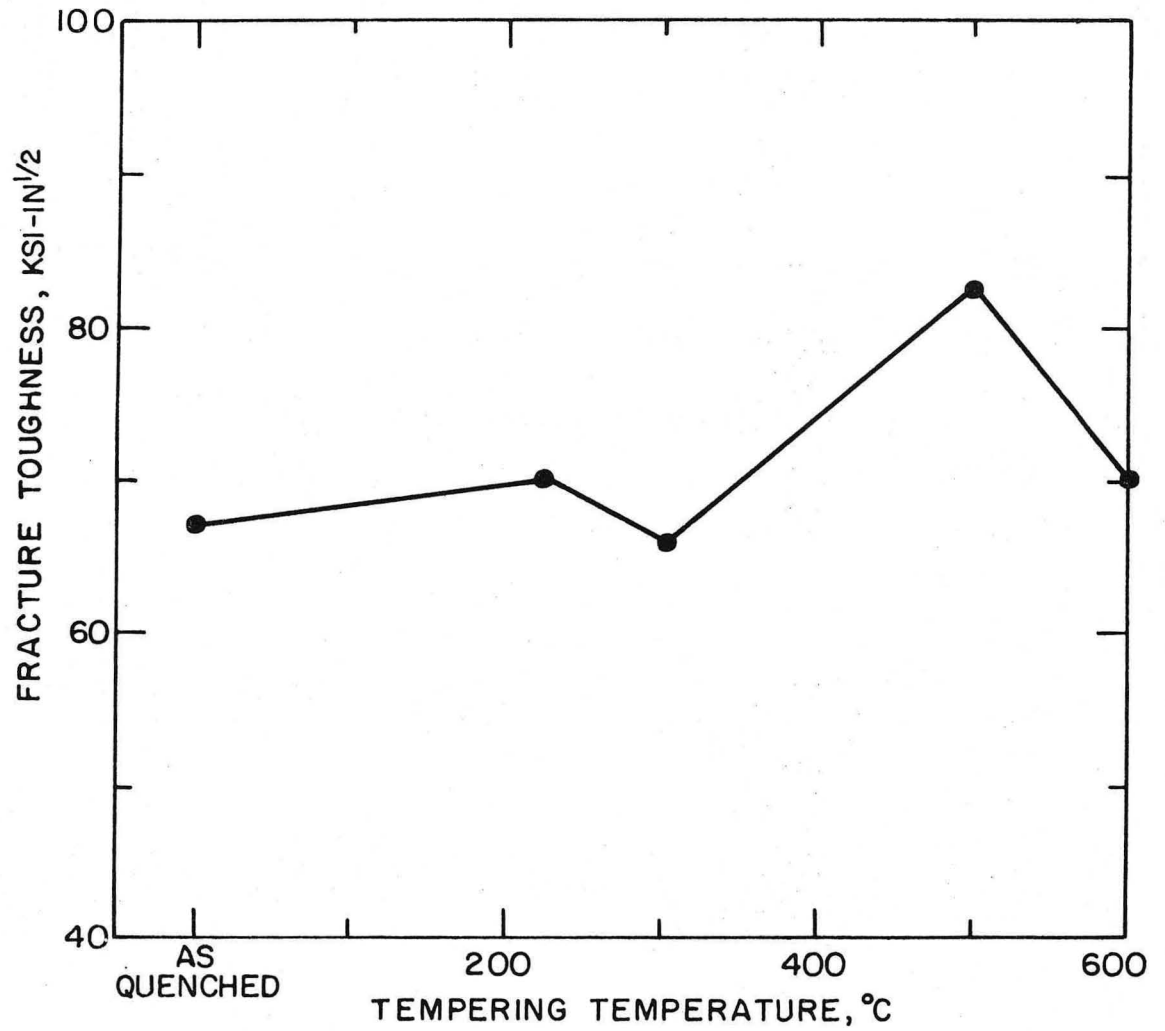


Fig. 26.



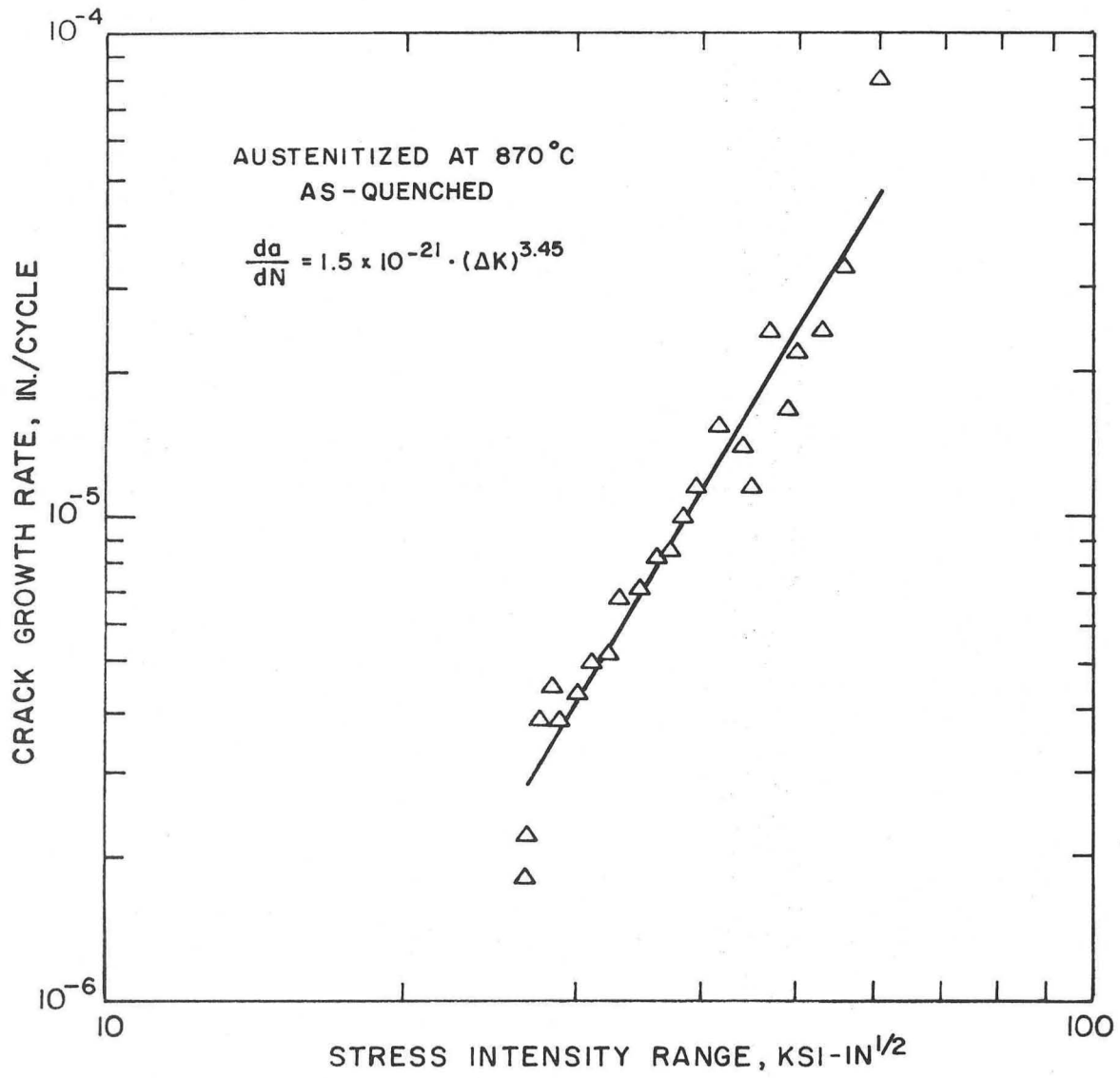
XBL73II-6603

Fig. 27.



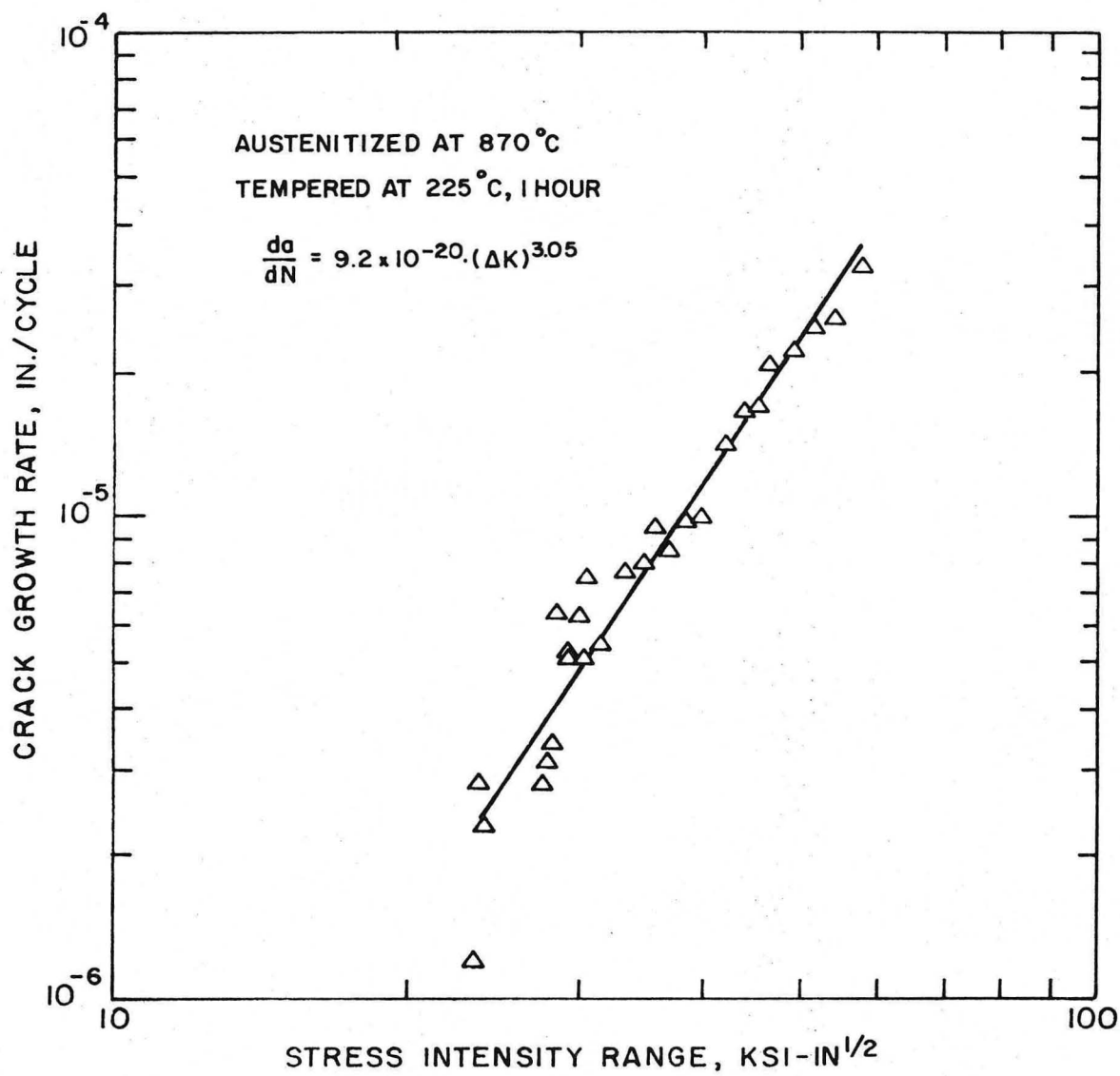
XBL73II-6602

Fig. 28.



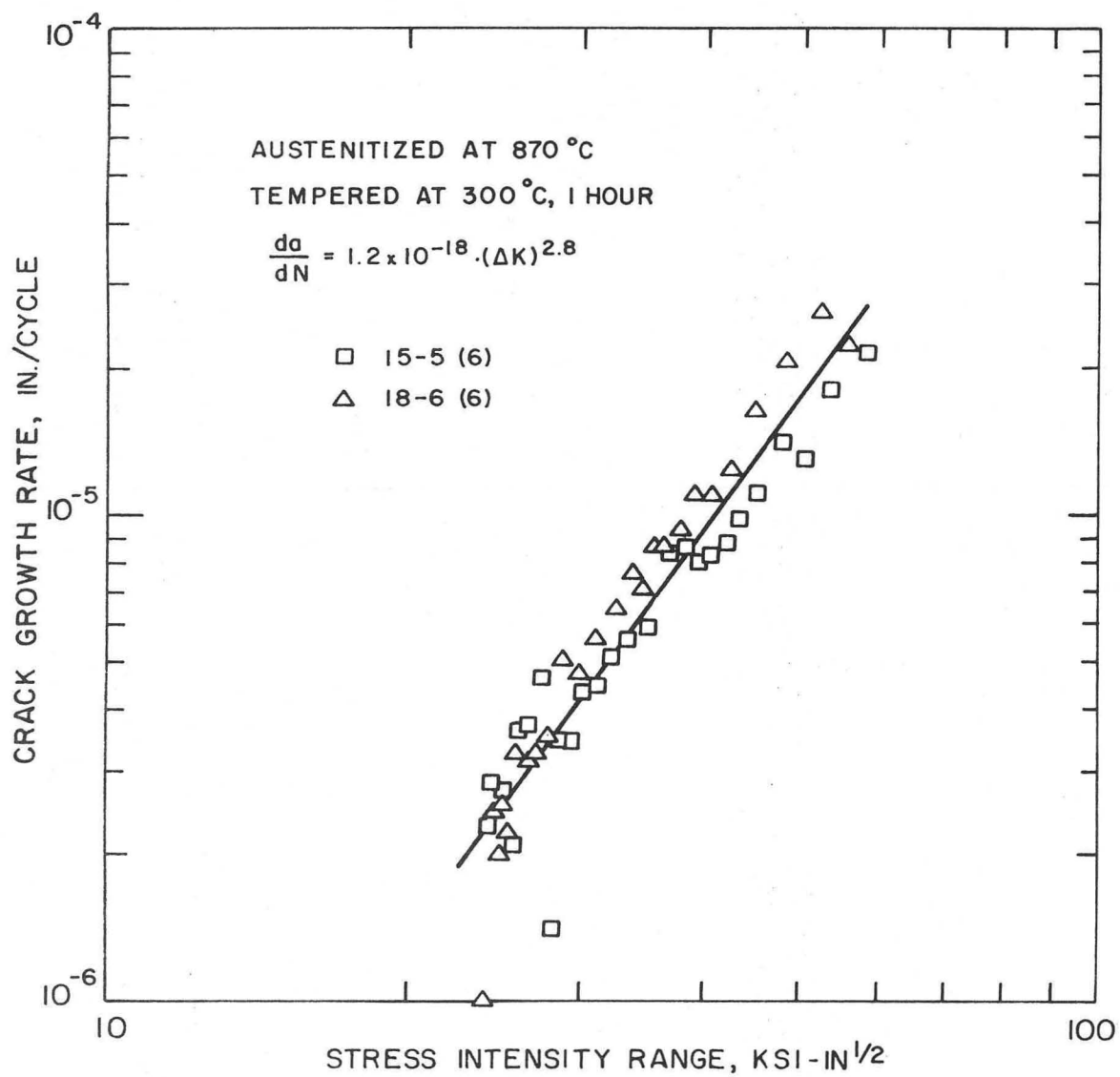
XBL7311-6608

Fig. 29.



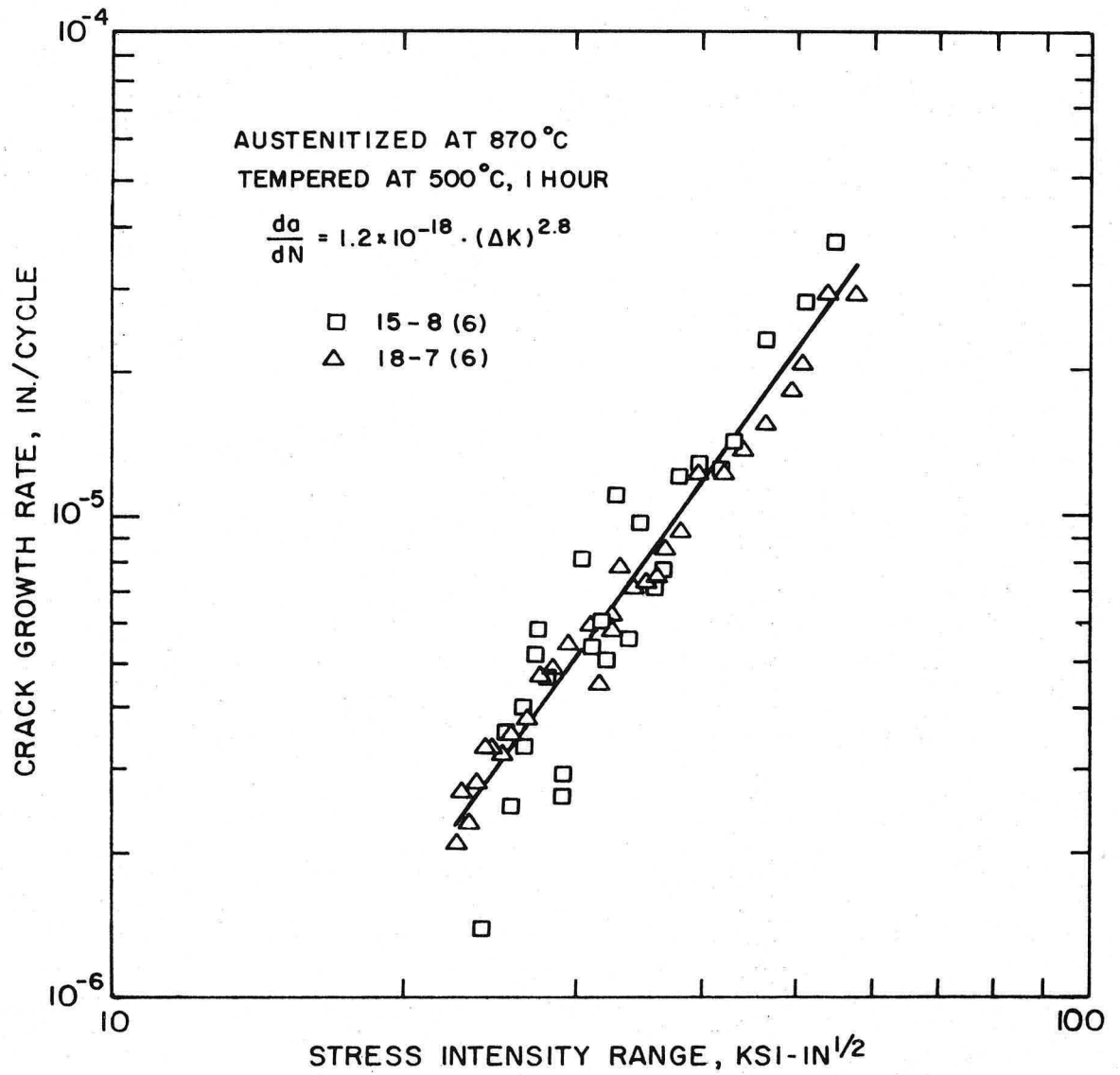
XBL 7311-6607

Fig. 30.



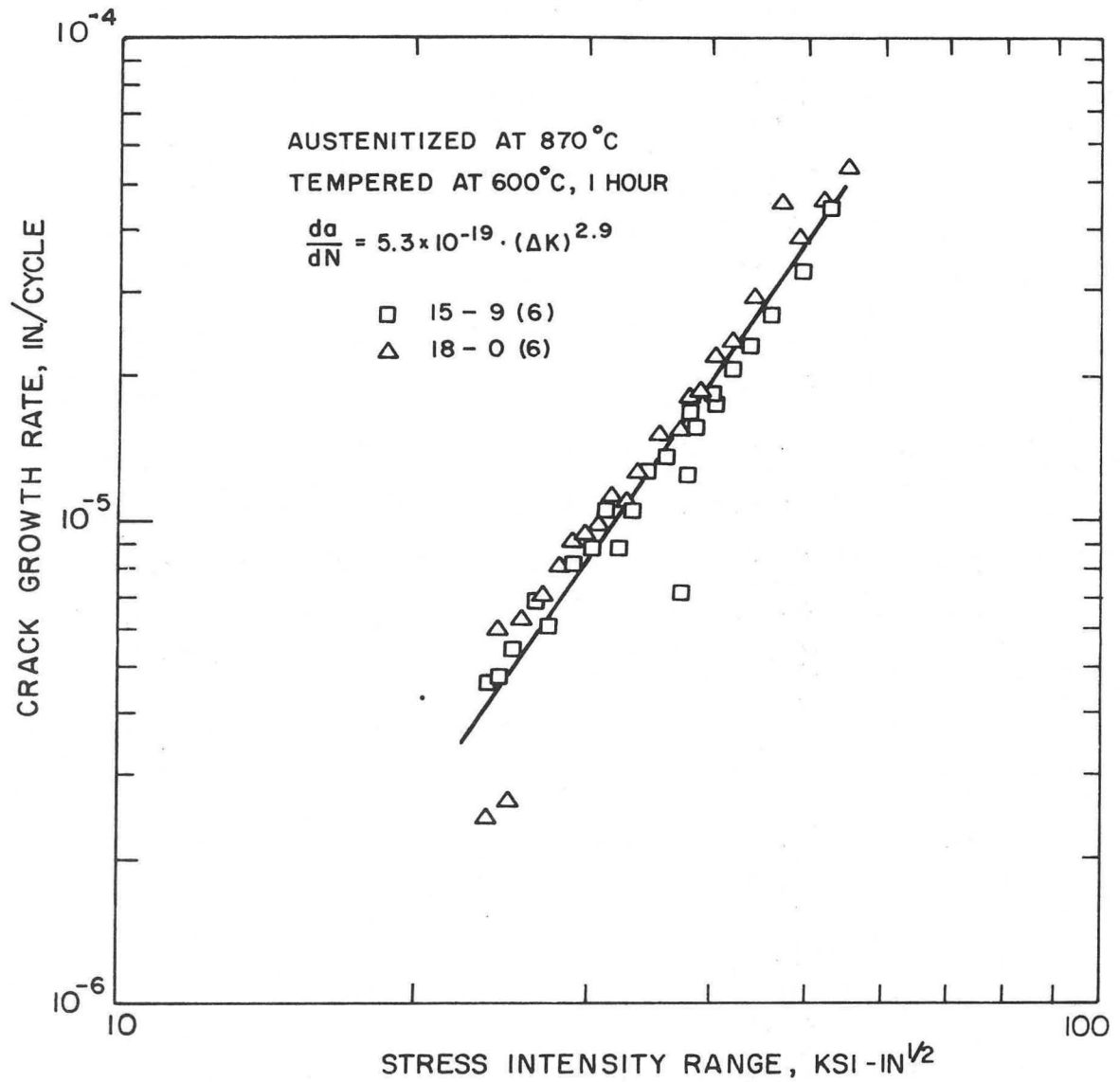
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Fig. 31.



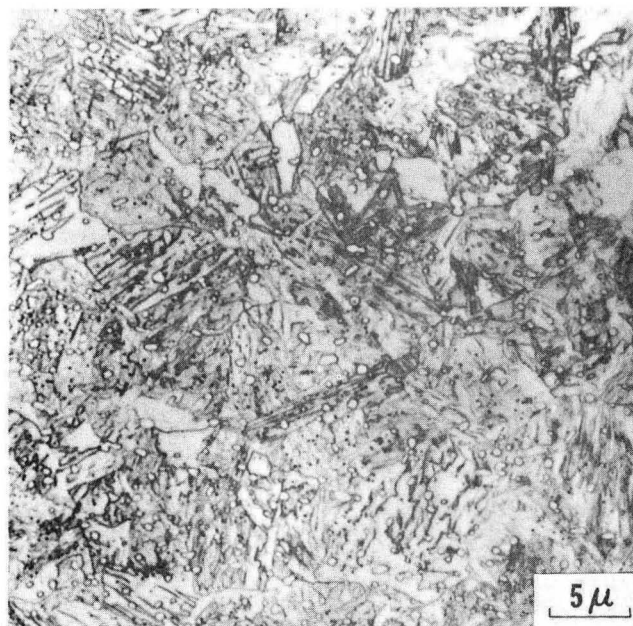
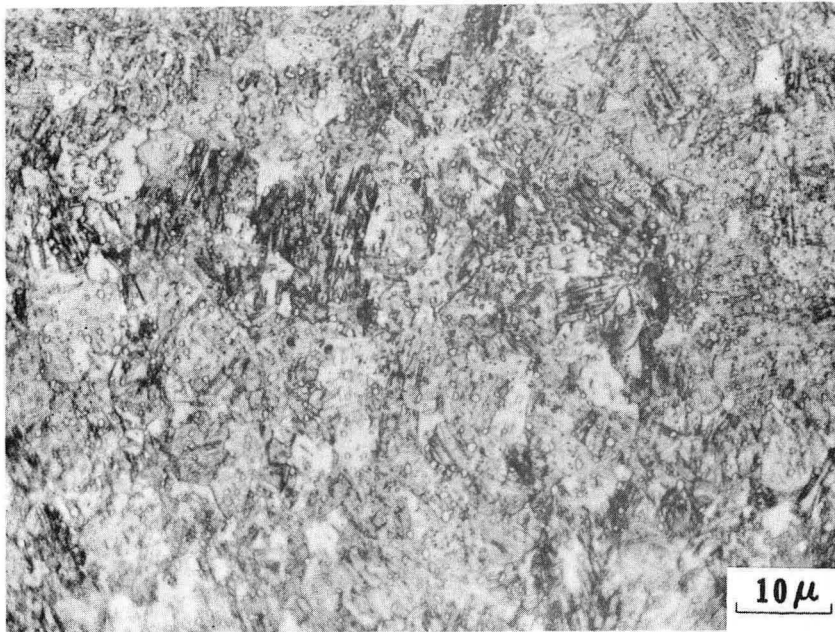
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Fig. 32.



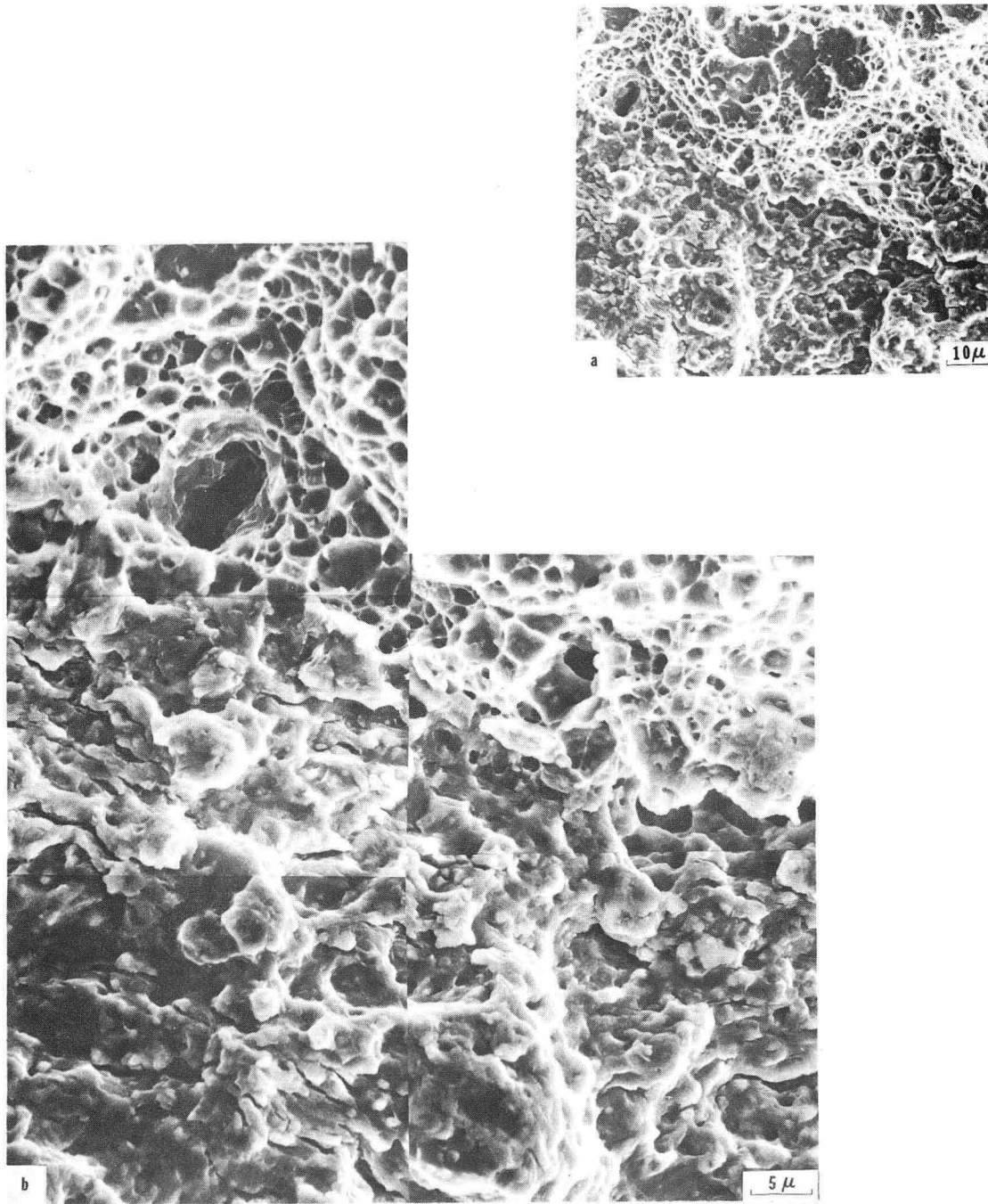
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Fig. 33.



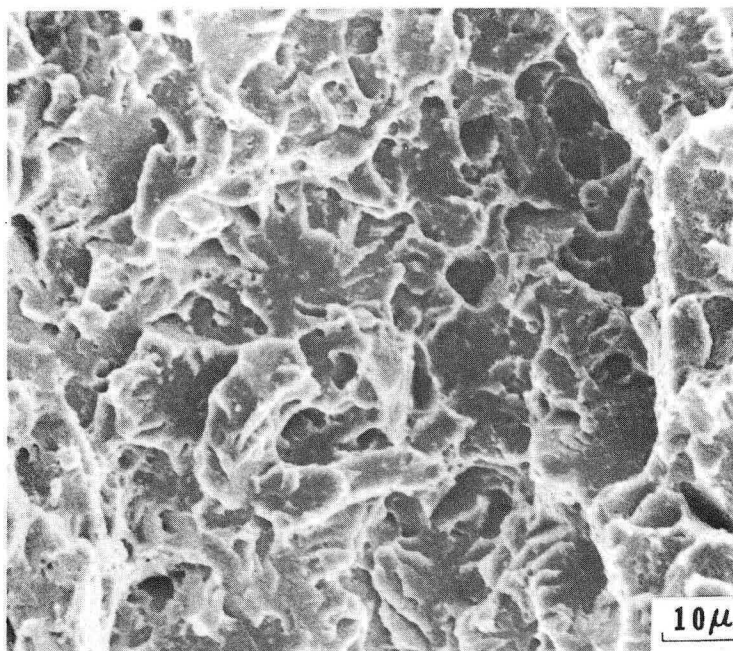
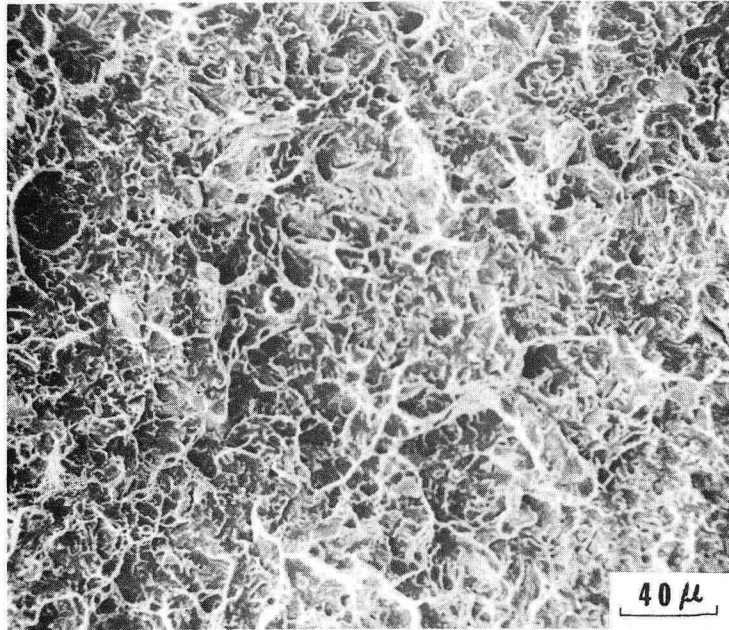
XBB 7311-6683

Fig. 34.



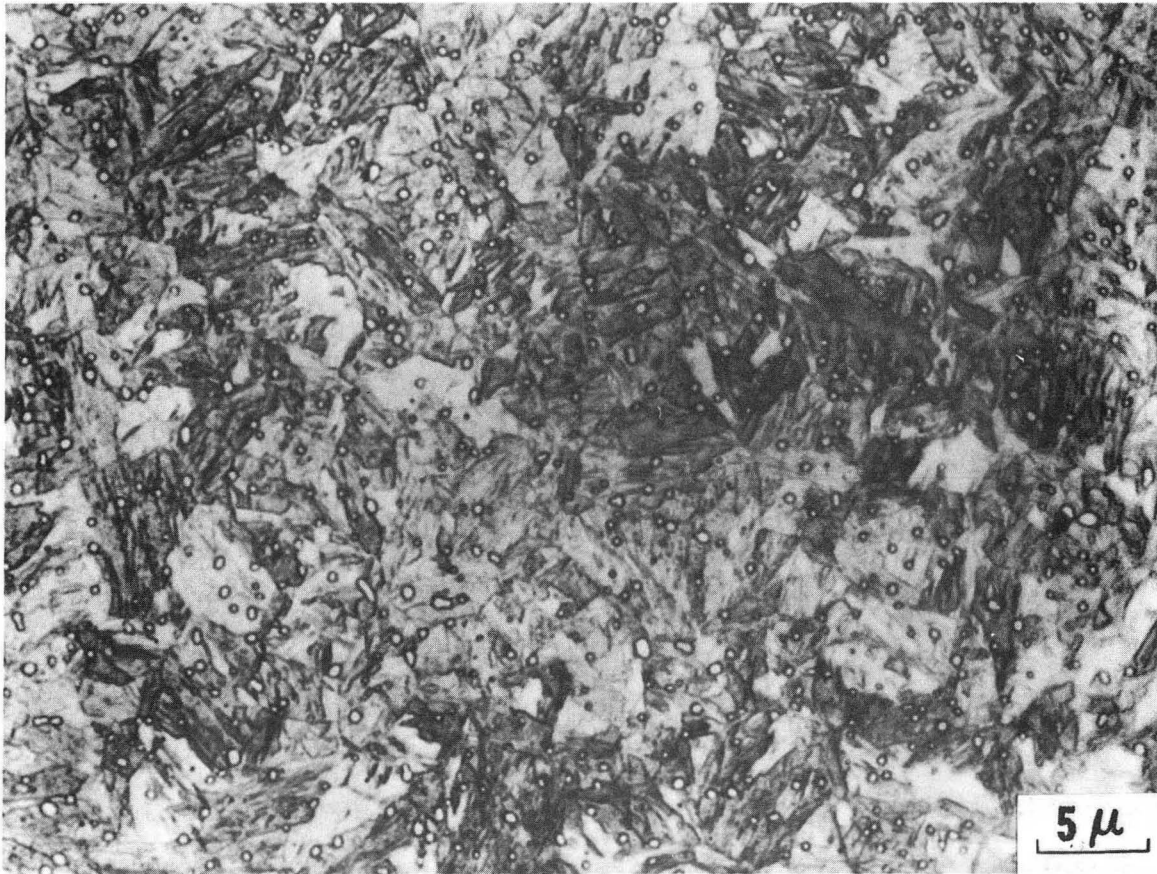
XBB 7311-6680

Fig. 35.



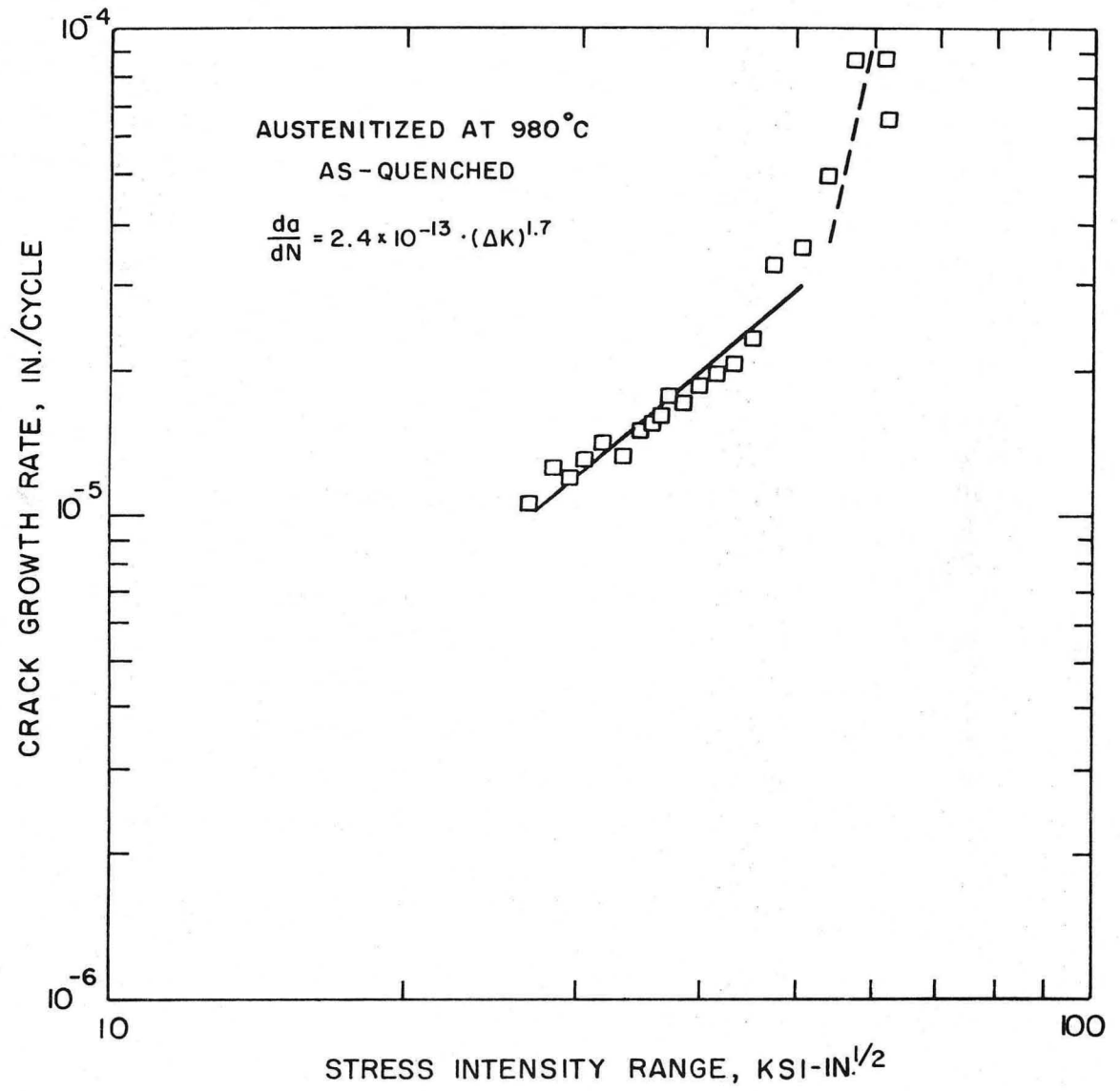
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Fig. 36.



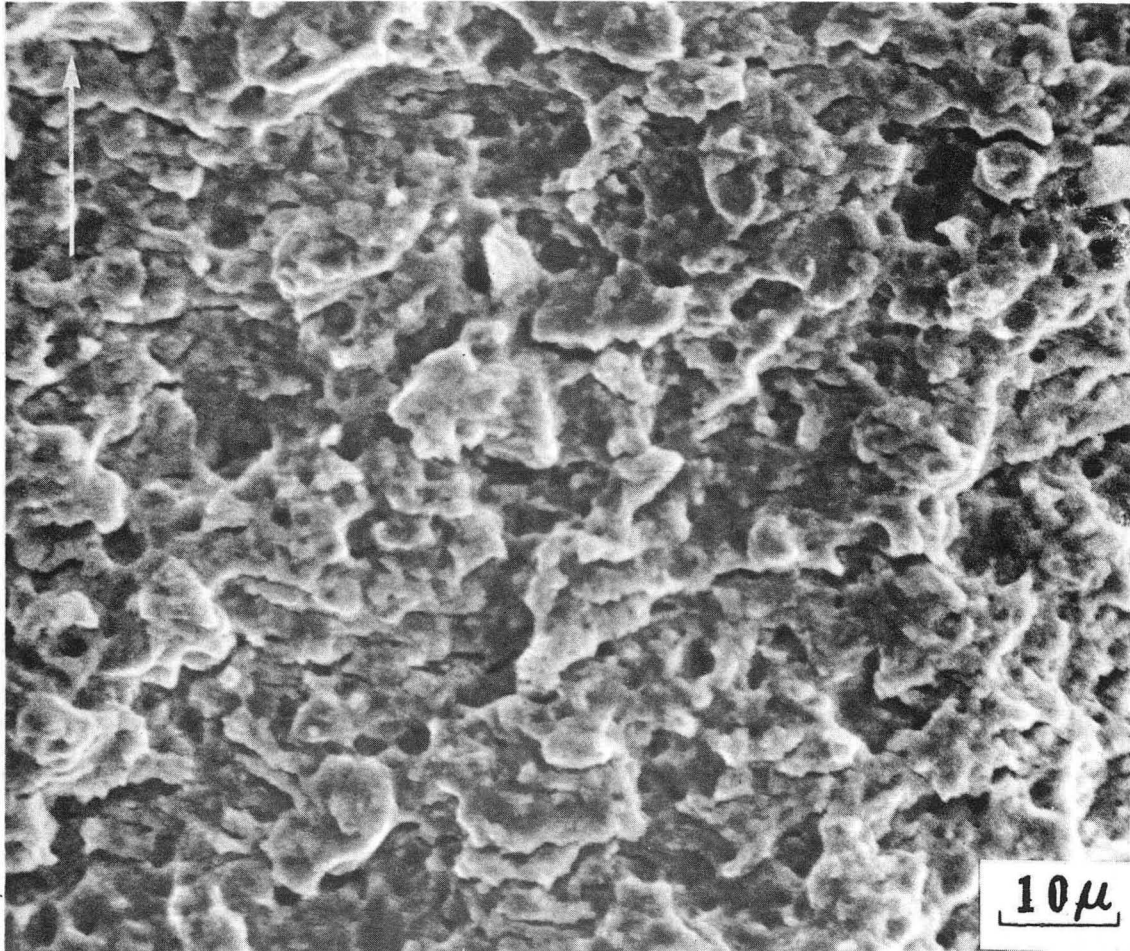
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Fig. 37.



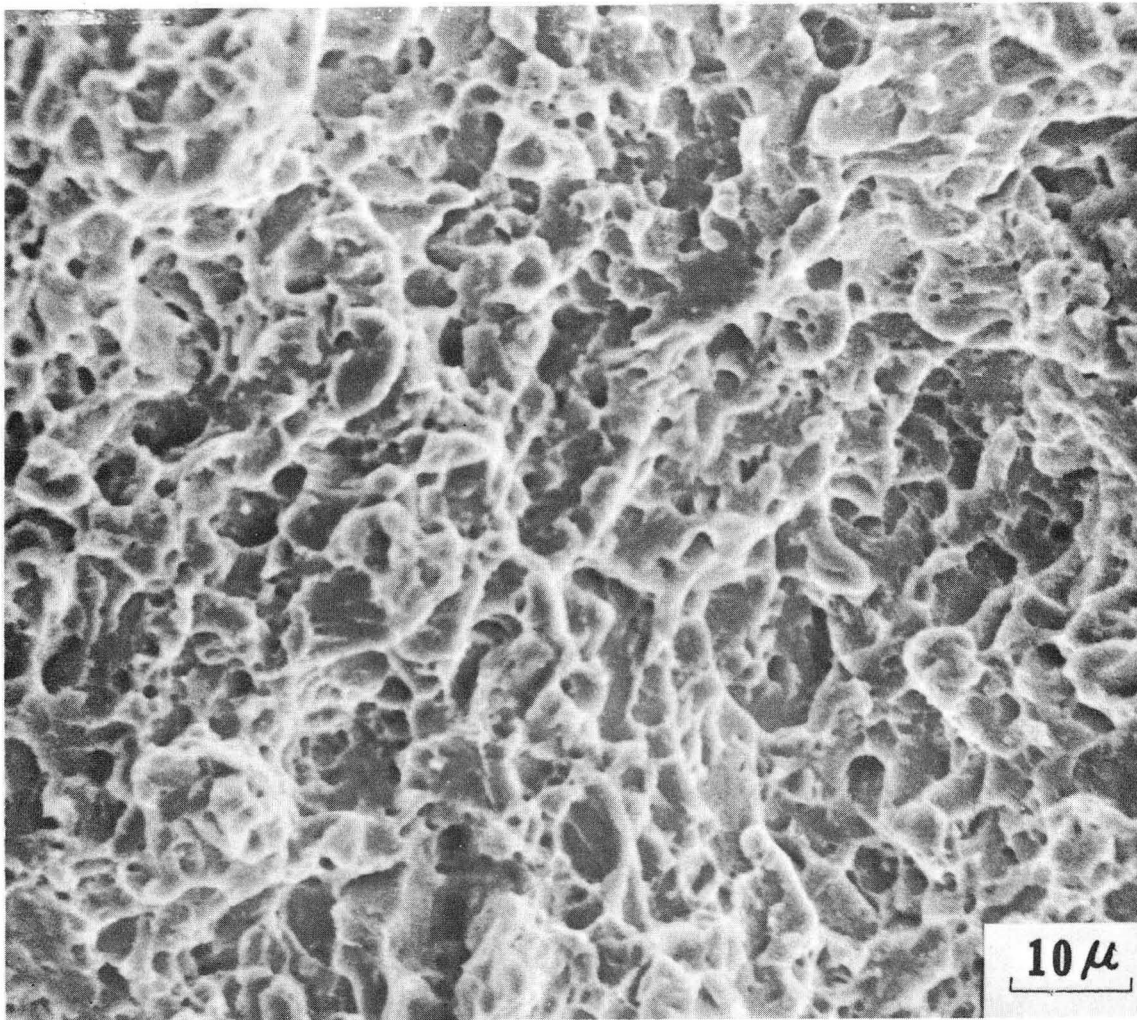
XBL7311-6611

Fig. 38.



XBB 7311-6693

Fig. 39.



XBB 7311-6684

Fig. 40.

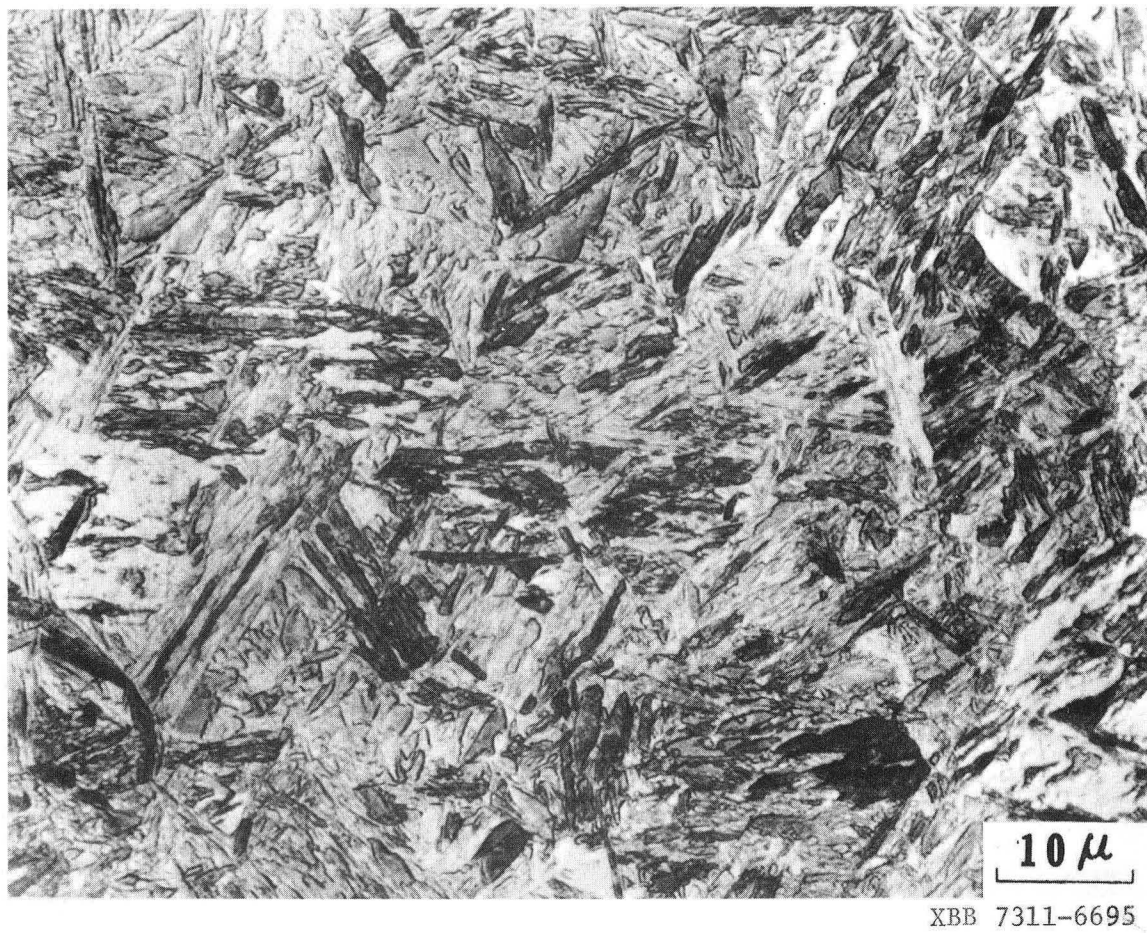
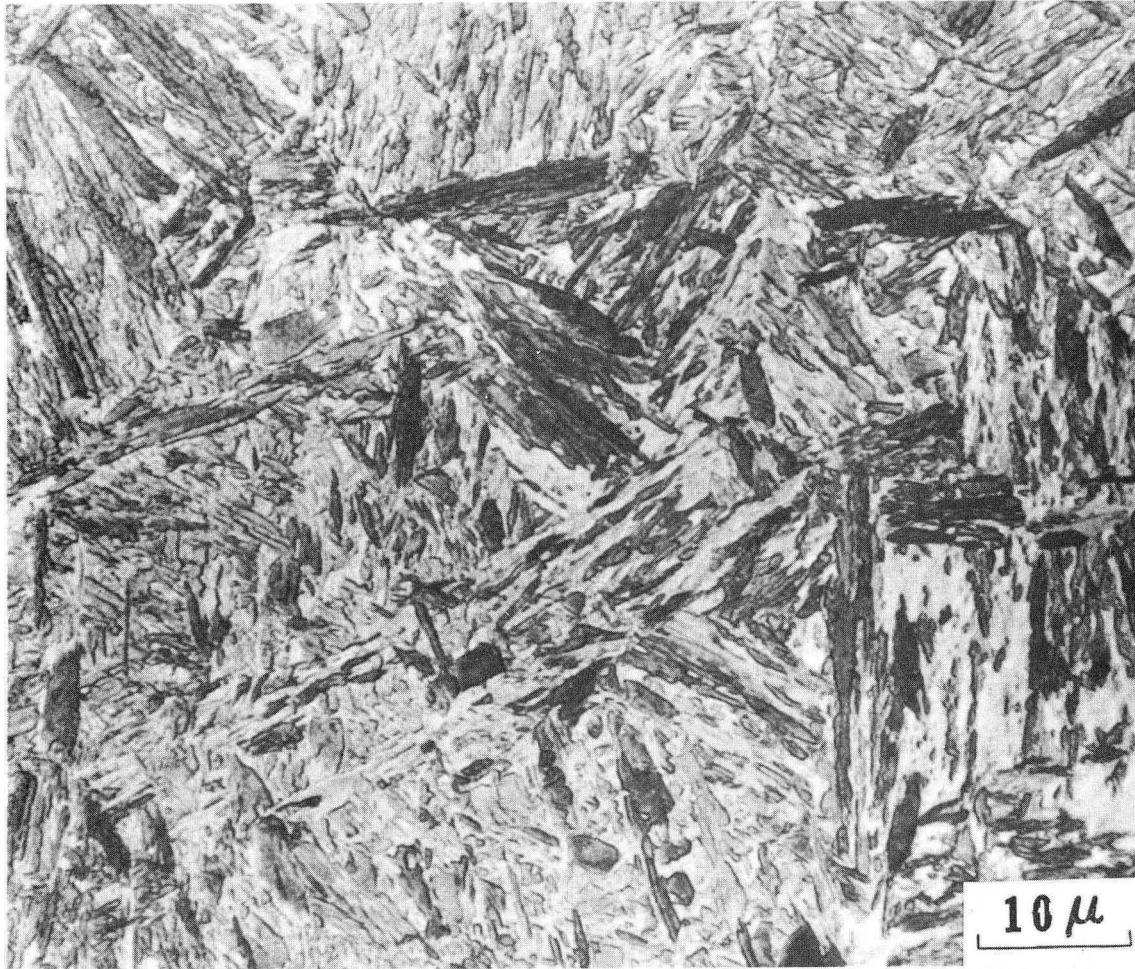
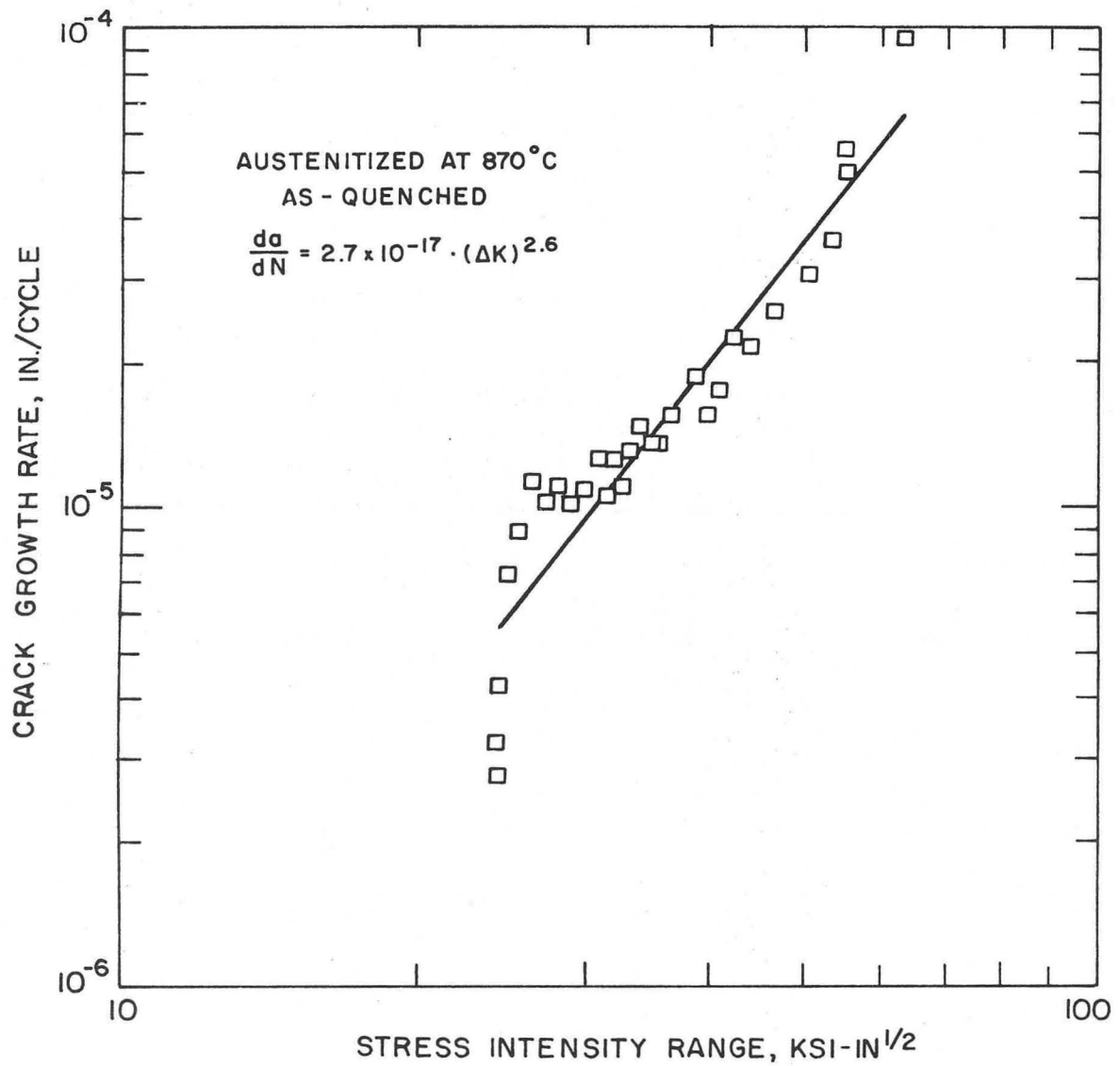


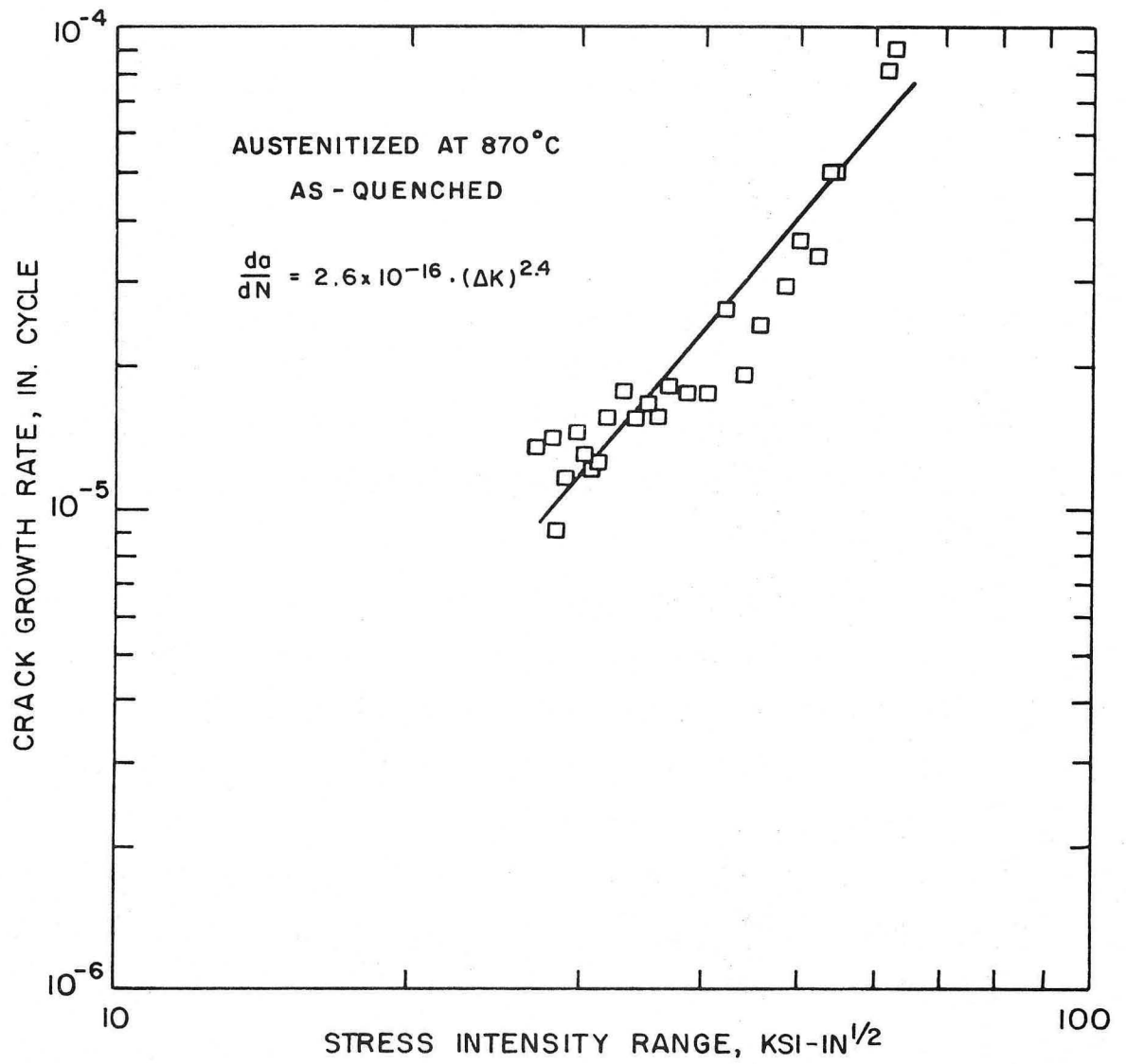
Fig. 41.



XBB 7311-6688

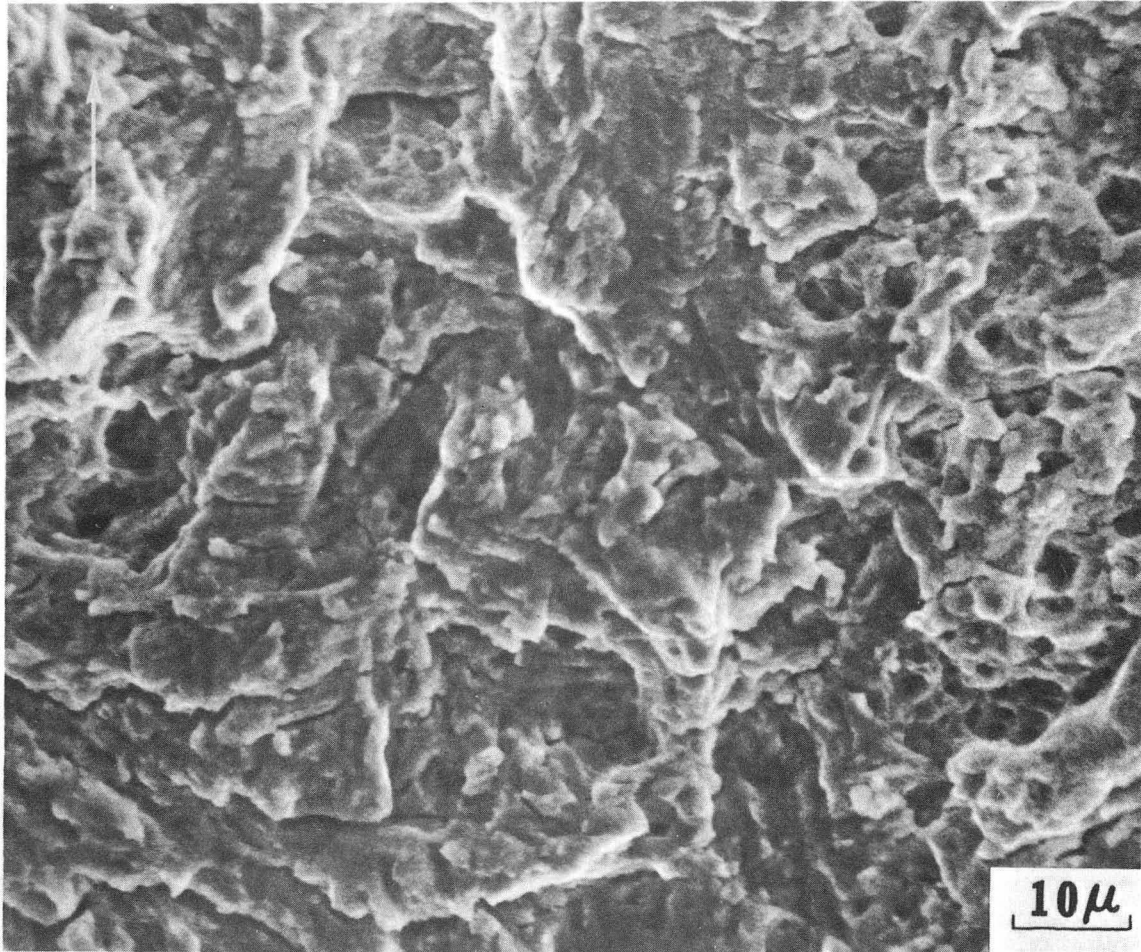
Fig. 42.





XBL7311-6610

Fig. 44.



XBB 7311-6690

Fig. 45.

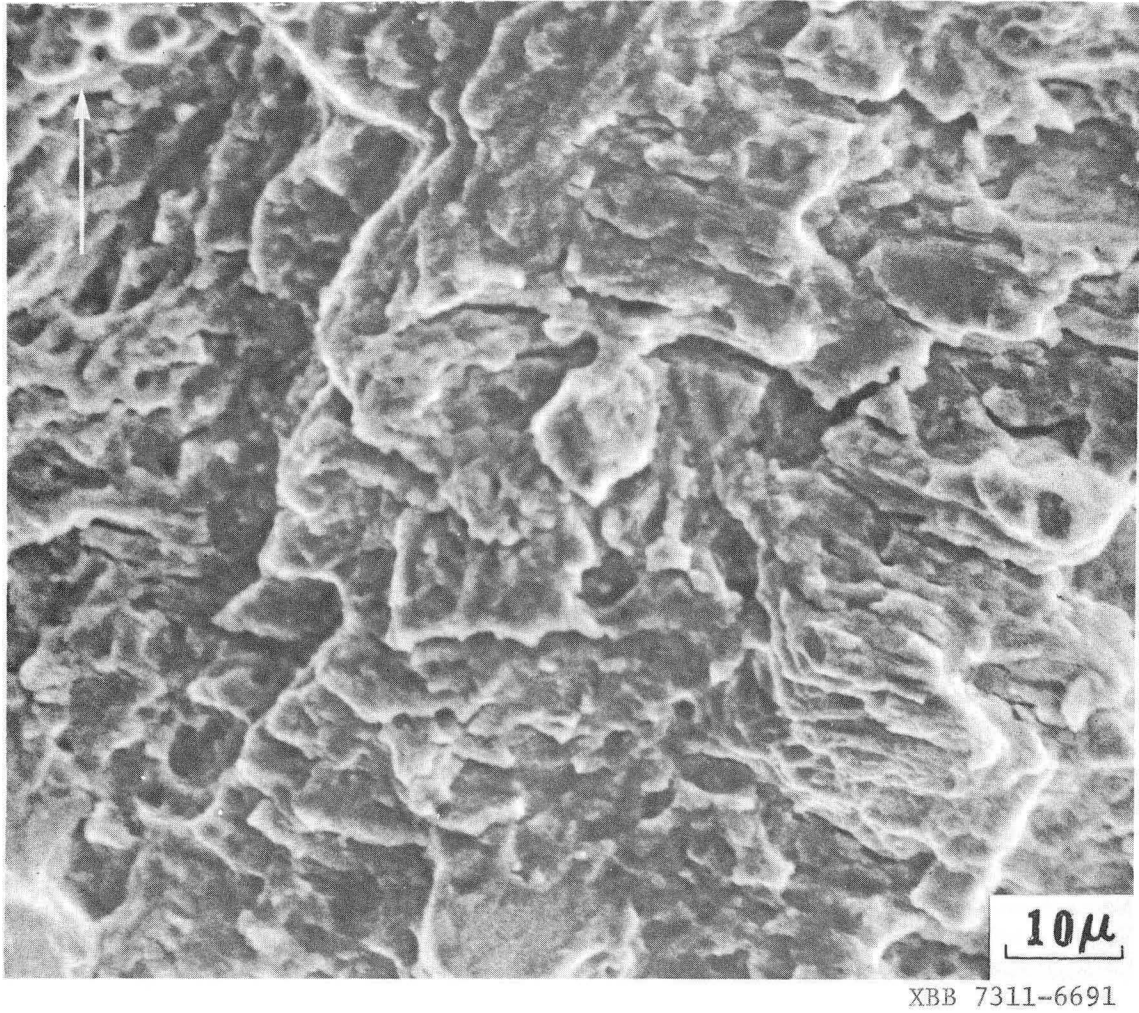
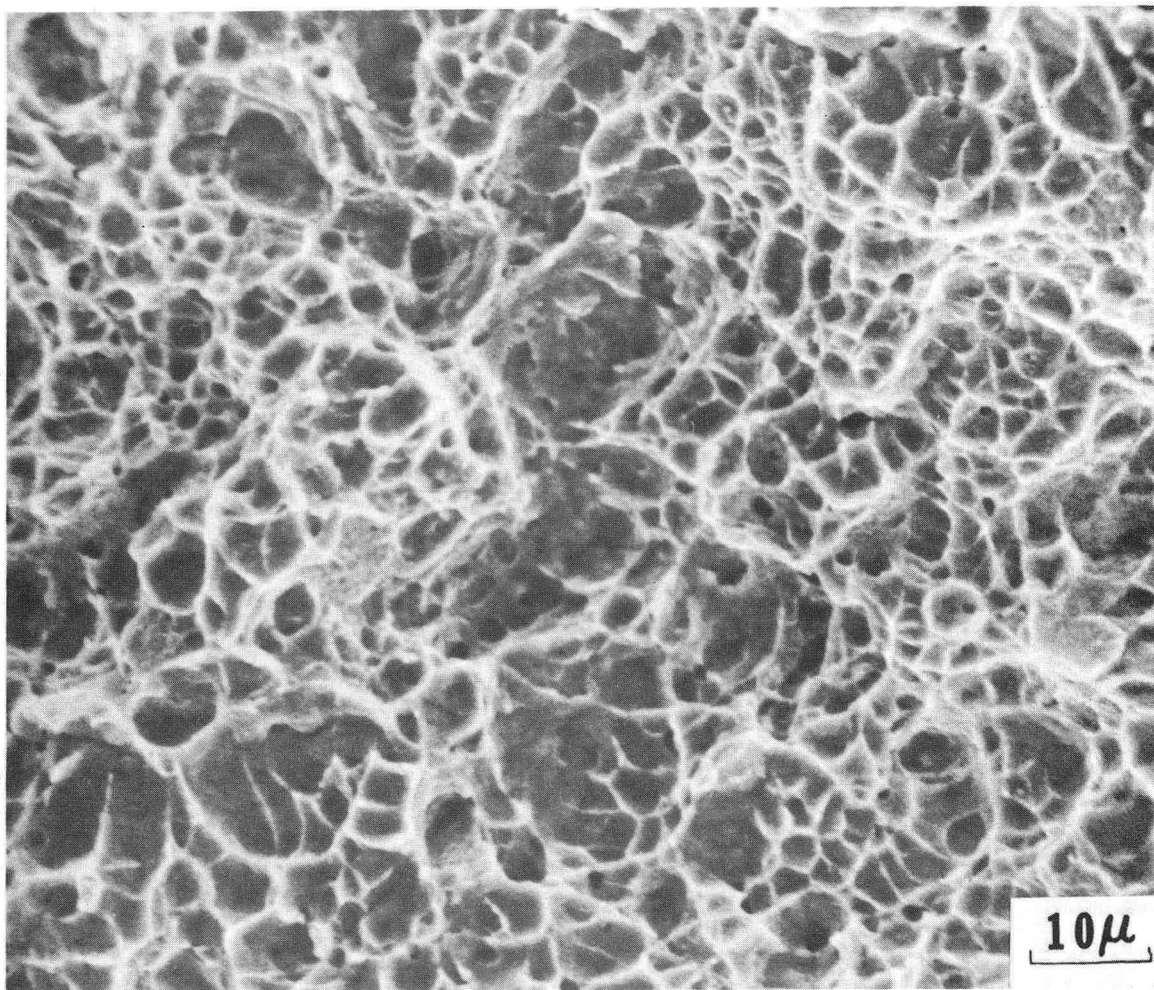
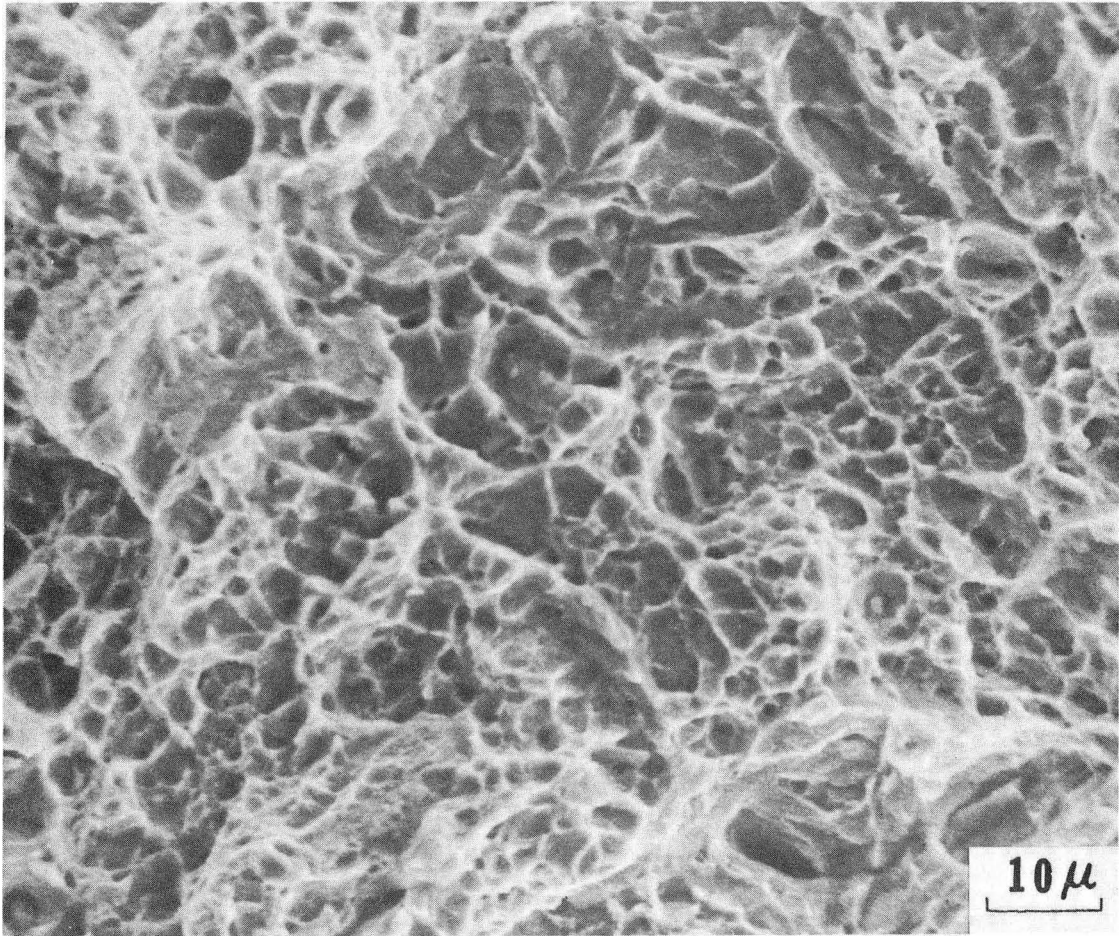


Fig. 46.



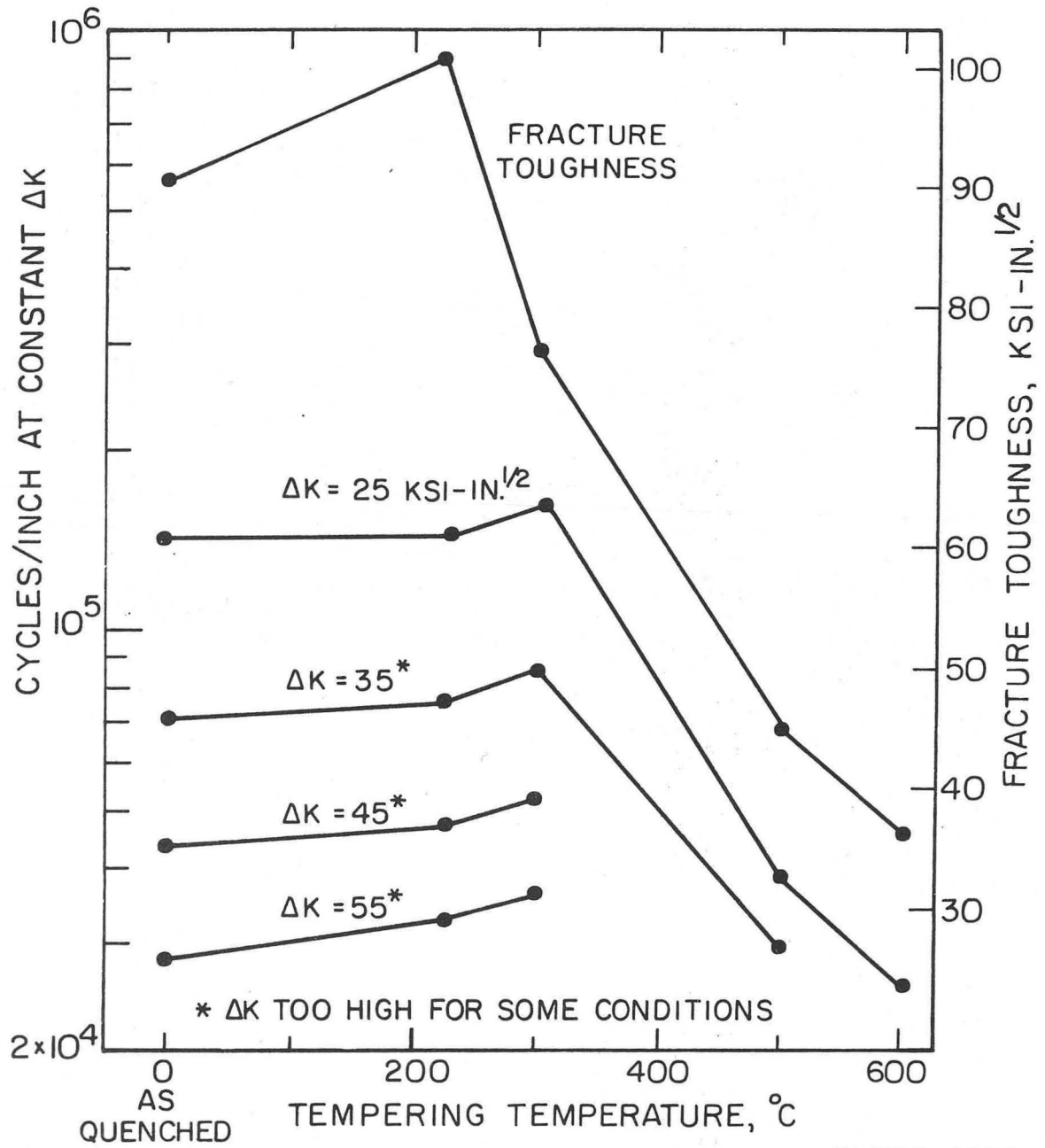
XBB 7311-6685

Fig. 47.



XBB 7311-6686

Fig. 48.



XBL 739-1860

FIG. 49.

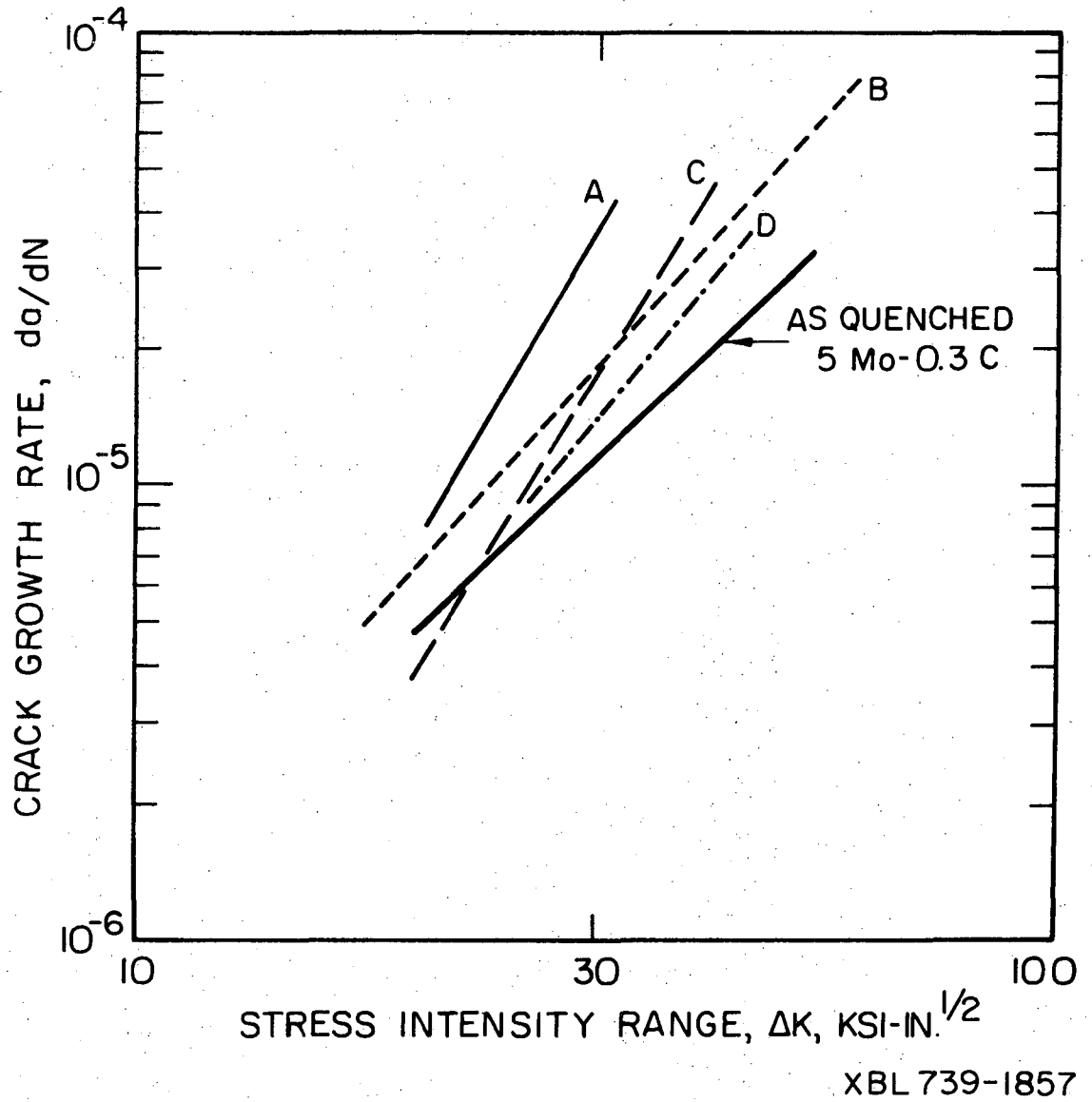
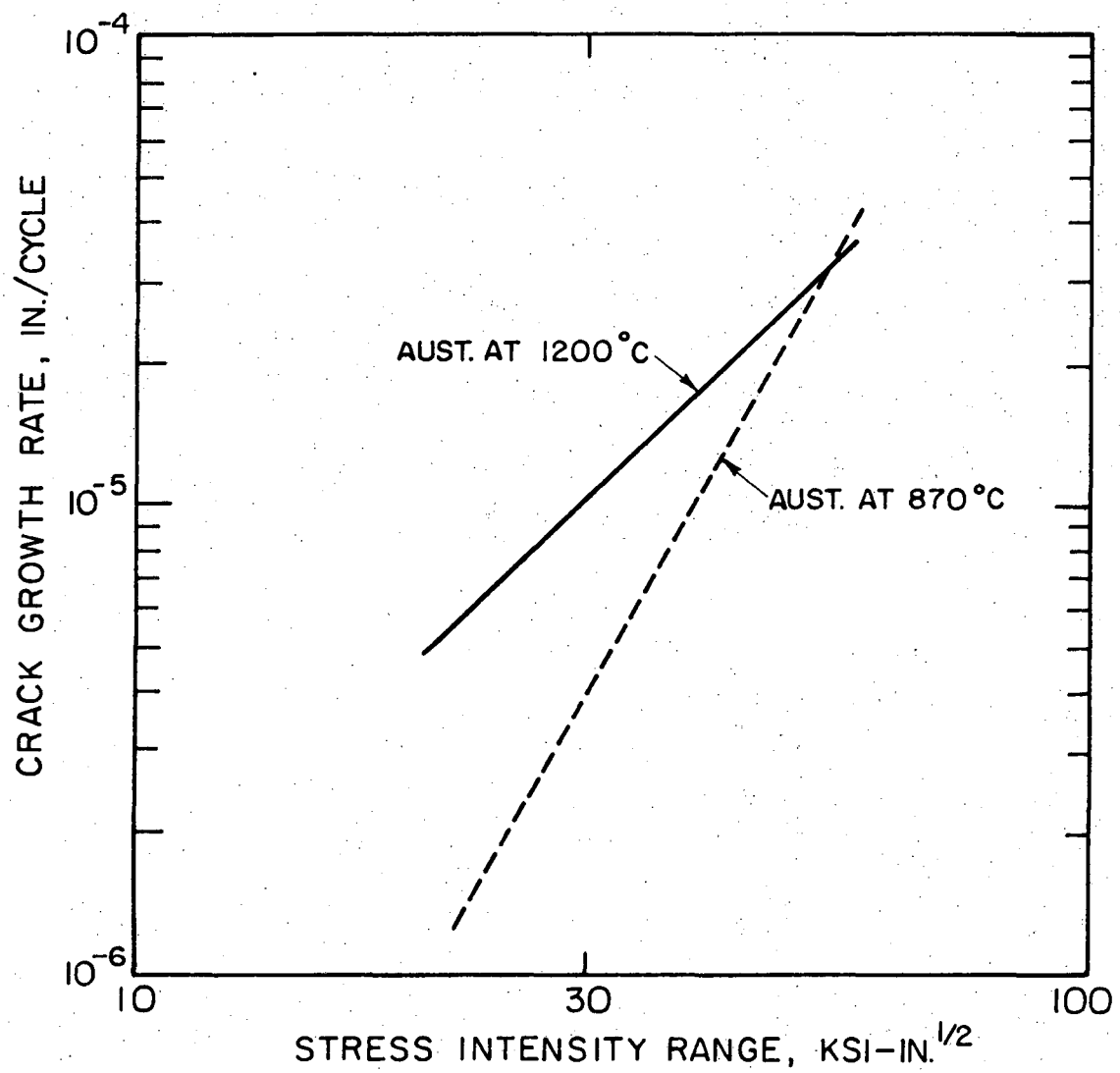
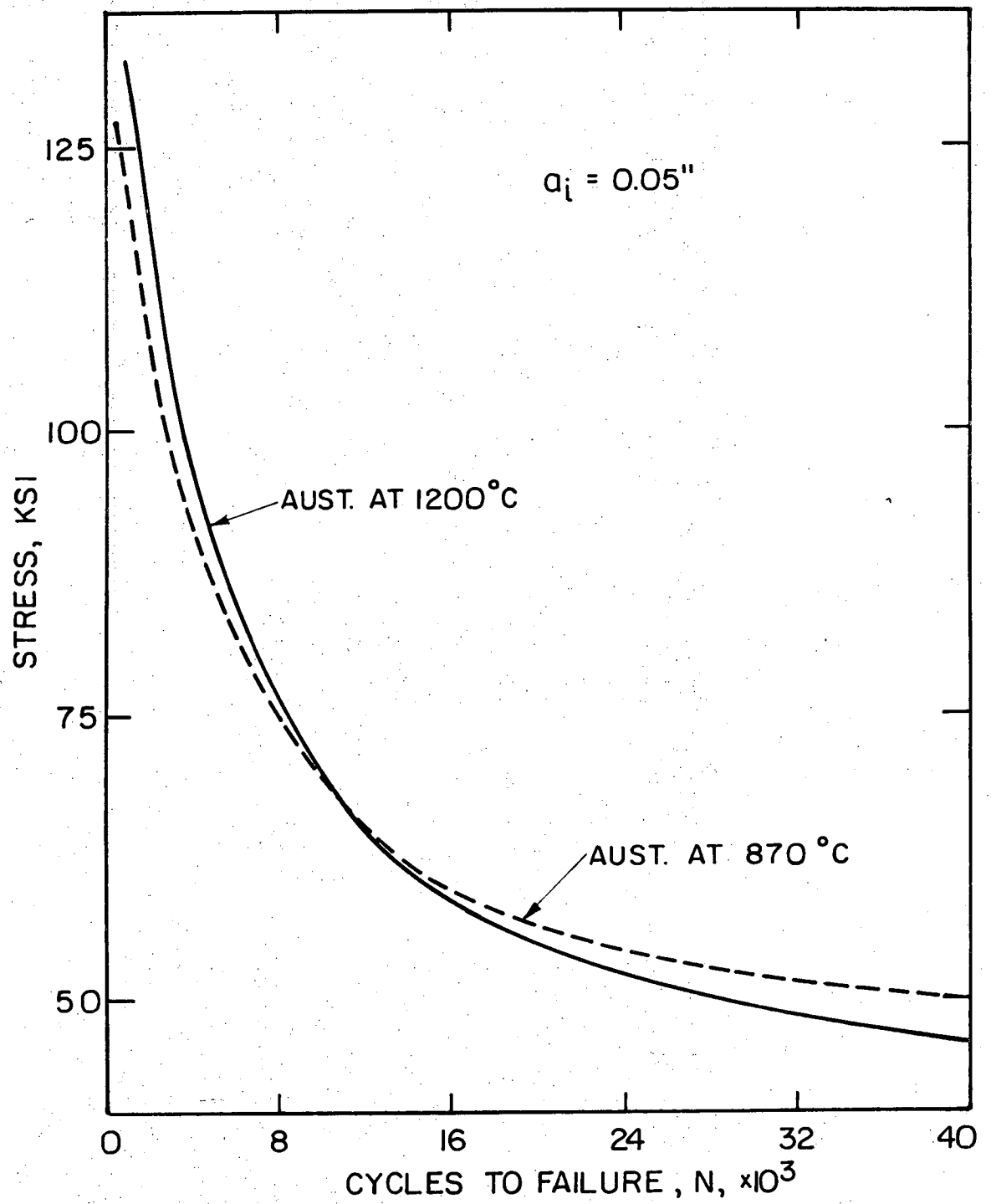


Fig. 50.



XBL 7311-6732

Fig. 51.



XBL7311-6733

Fig. 52.

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