THE MECHANISM OF CRACK INITIATION IN SILICON-IRON UNDER ELEVATED TEMPERATURE FATIGUE CONDITIONS

by

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ABSTRACT

The crack initiation mechanism in Fe-Si under elevated temperature conditions has been studied. Crack initiation was found to be not an easy process relative to crack propagation. The mode of failure was by production of intergranular cavities which linked together to form intergranular cracks. Cracks always initiated inside the sample. A model for nucleation of voids as a result of the grain boundary sliding at particles and ledges followed by the growth of voids due to the grain boundary sliding and/or vacancy diffusion and condensation at voids seems to agree with the experimental observations.

Thesis Supervisor: Arthur P.L. Turner
Title: Assistant Professor of Mechanical Engineering

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INTRODUCTION

Fatigue failure of metals has been a subject of interest since the turn of this century. It has been twenty eight years since Jenkins (1) observed one of the most controversial features of fatigue failure and named them "distortion cavities". Since then improvements in observation techniques have produced a large amount of information. Consequentally, new theoretical models have been introduced in the field of failure mechanisms of metals under elevated temperature fatigue conditions.

J. N. Greenwood et al. (2) observed inter-crystalline fracture in brass at elevated temperatures developing from small cavities in grain boundaries. They advanced a hypothesis to account for their observations, in which it was suggested that the cavities are formed by the agglomeration of vacancies. Machlin (3) suggested that generation of vacancies which is needed for vacancy condensation could occur due to plastic strain, while Balluffi and Seigle (4) pointed out the ambiguity of this and showed that sufficient vacancies can be produced at transverse grainboundaries because of the tension acting across the boundary. To Balluffi and Seigle it sounded more reasonable that there should be void formation during compressive strain if vacancies were generated due to plastic strain. However no void formation was observed in compressive creep, and this observation lead them to suggest the tensile stress as the cause of void formation.

D. Hull and D. E. Rimmer (5) did experiments under a combination of hydrostatic pressure and uniaxial tension. They proposed a relation between applied stress and the critical radius of voids beyond which cavity growth can continue indefinitely. However, the experimental relationship between strain rate and applied stress obtained by Sherby (6) showed a discrepancy with this formulation. Further serious objections were raised against the above theoretical model when Boettner and Robertson (7) measured the density change due to void growth. In reply to these objections Greenwood (8) was able to account for these discrepancies by assuming a constant rate of nucleation of voids.

The idea of vacancy diffusion was adapted by R. P. Skelton (9), and he suggested that the grainboundary self diffusion varies with the applied stress and that the lattice vacancies reach the cavity surface from a "life-time" sphere around the cavity. Broomfield (10) disagreed with Skelton and claimed that the main flux of vacancies comes from the grainboundary. Gittins (11) claimed that the assumption of diffusion of vacancies to spherical cavity nuclei situated on stationary grainboundaries should be modified to include defects absorbed by grainboundaries during migration which contribute to cavity growth.

A. S. Argon's work on lithum floride (12) and R. Resnick's work on brass (13) belong to those which agree with the vacancy diffusion model. There are many observations like those made by T. Ogura and S. Karashima (14) of void formation near fatigue cracks

in 99.99% pure copper.

There have been arguments against the vacancy diffusion mechanism since its first introduction by Greenwood (2). R. D. Gifkins (15) presented an alternative hypothesis based on the formation of voids at jogs by grainboundary sliding and argued that his model also fits the experimental observations made by Greenwood quite well. At the same time, C. W. Chen and E. S. Machlin (16) did experiments on copper bicrystals of 99.999% purity and gave supporting evidence for Giffin's suggestion and made some modifications to his theory. Thereafter, nucleation of voids at ledges on a boundary as a result of grainboundary sliding has been discussed by many people, and McLean (17) claimed that grainboundary ledges only four atoms high should be able to nucleate cavities.

K. U. Snowden's work (18) on bicrystals of lead showed the possibility that grainboundary sliding is responsible for the nucleation of intercrystalline cracks. Meanwhile, H. D. Williams (19) modified the idea of grain boundary sliding such that grainboundary sliding at particles causes the decohesion of particle-from the matrix and that these sites are good places for void nucleation. H. D. Williams (20) also tried to measure the angular distribution of cavitated boundaries using the statistical technique of Scriven and Williams (21), and showed that the incidence of cavitation is maximum around the plane of maximum shear stress, thus supporting the sliding argument. The angular distribution of

cavities seems to be different in fatigue conditions and in creep conditions. R. V. Day (22) and A. L. Wingrove (23) showed that under creep conditions maximum cavitation occurs around 60°--70° in contrast to William's measurements in fatigue conditions.

There appears to be a trend among investigators to favor a combination of vacancy diffusion and grain boundary sliding as the mechanism of fatigue. In some cases, grain boundary sliding is preferred as a nucleation mechanism and vacancy diffusion as a growth mechanism. Other mechanisms of crack initiation like oxidation of grainboundaries which slide out of the surface as suggested by J. Hannoosh (24) could also be considered.

As shown in the above brief review many mechanisms of crack initiation have already been suggested which have their own supporting evidence. Like any other problems, there are a large number of variables that must be considered when the question of a fatigue mechanism arises. No one really could answer this question for all possible combinations of stress systems, sample conditions, temperature and environment. This work is limited to the question of crack initiation mechanism of Fe - 3%Si under vacuum and elevated temperature, above half of the absolute melting temperature.

A previous study (24) indicated that cavitation was doubtful and left the mechanism of crack initiation unexplained. This work was undertaken to investigate several of these unanswered questions.

Further more improved techniques for examining the sample allowed the work to correct several misinterpretations which occurred in the earlier work.

EXPERIMENTAL

A) Samples and Material

The material was iron - silicon 3%. Its chemical analysis is given in Table I. The geometry of the sample is shown in Fig. 1.

Two types of heat treatment were used to obtain two different grain sizes. Samples were annealed in vacuum for 8 hours at 1000° C to obtain a grain size having a mean grain diameter of 3×10^{-2} in. Other samples were annealed in vacuum for 4 hours at 1200° C. This annealing treatment gave a grain size of .18 in. mean diameter.

B) Testing System

The experimental system was composed of a constant deflection amplitude fatigue machine with a vacuum jacket containing a furnace internal to the vacuum jacket. A schematic diagram of the system is provided in Fig. 2.

A Tatnall-Krouse constant amplitude fatigue machine was used. The frequency of the machine was fixed at 1000 cycles per minute so that frequency effect was not studied. The vertical displacement of the eccentric cam was transmitted through an adjustable vertical rod to one end of the sample. The other end of the sample was fastened to the grip so that the sample was loaded as a cantilever beam. Tests were run at constant displacement amplitude measured at the moving end of the beam.

A pyrex tube was used for vacuum jacket. Vacuum was maintained by a fore pump and an oil diffusion pump.

The temperature system consisted of a resistance furnace inside the vacuum system. The temperature was controlled by a cycling type temperature controller using a sensing thermocouple near the furnace coils. The specimen temperature was monitored by a separate thermocouple.

C) Tests and Observations

From the data obtained for Fe - Si by another investigator (24), the relationship between displacement amplitude and the expected fatigue life was known. Two deflection amplitudes were used. One was .120 in. which corresponded to a life of 10^5 cycles. The other was .060 in. which corresponded to a life of 2×10^4 cycles. All tests were performed at 900° C.

Some samples were run with interruptions at 1%, 2%, 4%, 8%, 16%, 32% of their expected fatigue life. At every interruption the sample was observed and photographed using a stereoscopic optical microscope. Thus a continuous record of the development of the fatigue damage as far as it was visible on the outside of the sample was obtained.

Some samples were run for a certain percentage of their expected fatigue lives, and sectioned. A jeweler's saw was used to cut the section parallel to the stress axis. The sectioned specimen was polished successively with 80, 120, 240, 320 and 600

grit abrasive paper sufficiently to remove the damage introduced by sawing and then polished with aluminum oxide of 0.3 micron grain size and finally with aluminum oxide of 0.05 micron grain size.

Chemical etching was used to remove the cold worked layer and to reveal the grain boundaries and the grain structure. A 5% nital etch which is composed of five volume percent of 70% nitric acid and ninety-five volume percent of methyl alcohol was used as the etching solution. The etching period used was 20 sec. After etching the sample was rinsed in alcohol and then cleaned in an ultra-sonic cleaner to remove all traces of the etching solution from cracks in the sample prior to drying.

Even with ultrasonic cleaning it was difficult to remove all of the chemicals from the cracks in the sample so that it would dry without staining. The chemical agents used penetrated into the cracks and had not dried out even though the surface appeared to be quite dry. The remaining chemical solution, mostly alcohol, later came out of cracks and made colorful spots along crack sites. Although this confirmed the existence of fine deep cracks along the grain boundaries, it was inconvenient for detailed observation of the sample. The problem was solved by using petroleum ether as a final rinse after the ultrasonic cleaning. The high volatility of petroleum ether helped the drying process making it possible to dry the solution out of the cracks without staining.

After the samples were prepared, observations were made with an optical microscope and with a scanning electron microscope.

Table 1--Analysis of Silicon Iron

Element	Weight %
Carbon	.03
Manganese	.058
Sulphur	.019
Silicon	3.29

Table 2--Mechanical Properties of Fe-Si

Temperature; °C	Y; psi	E; psi	ε_e
630 (1/2 Abs.T_m)	15×10^3	12×10^6	1.25×10^{-3}
900 (2/3 Abs.T_m)	2×10^3	6×10^6	3×10^{-4}

RESULTS

A) Development of Fatigue Damage on the External Surface

Observations were made on the samples just after they were cycled to a certain fraction of their expected fatigue life, for example after 1%, 2%, 4%, 8%, 16%, 32% and 100% of life, to determine what damage had occurred and if cracks had been initiated at the surface. Pictures were also taken at these times.

a. Small Grain Size Samples at .060 Displacement Amplitude

At 1% of fatigue life 1000 cycles, grain boundary sliding was observed. There was no obvious sign of deformation internal to the grains except very near the boundaries as shown in Fig. 3a. At 2% of life the magnitudes of the permanent offsets at grainboundaries were observed to have increased. The presence of furrowed regions along the boundaries indicated that grain boundary migration had occurred. After 4% of life, both permanent grainboundary sliding and grainboundary migration were intensified. Some grains were observed to have been pushed out of the surface by distances approaching 25% of a grain diameter as shown in Fig. 3c. However, the interior of each grain was still free from obvious signs of plastic deformation.

Until this stage of fatigue damage, there had been very little fatigue damage along the center plane of the fatigue specimen which was the neutral axis of the beam as shown. By 8% of life the surface had become rougher as shown in Fig. 3d . Slip lines and curvature of the surfaces indicated the beginning of

of extensive plastic deformation inside the grains. Both permanent grain boundary offsets and grain boundary migration had become very extensive by this time. No features which could definitely be called cracks were observed at this stage, but sharp notches caused by grain boundaries sliding in the direction out of the surface plane were common. Damage was observed even along the neutral axis of the sample as shown in Fig. 4a. At 16% of fatigue life, damage including grain boundary sliding and grain boundary migration began to be confined to certain regions of the sample near the minimum gauge section, as shown in Fig. 3e. Extensive surface roughening by both plastic deformation and grain boundary sliding had occurred by this time, but still no crack initiation was observed on the surface. As will be discussed later, cracking inside the sample seemed to have begun by this stage.

At 32% of life surface deformation was so severe around the minimum section that it was hard to distinguish the grain boundaries (Fig.3f). Surface cracks were finally observed. The region of extensive deformation had grown since 16% of life. Observations of the broken sample at the completion of the test revealed that severe plastic deformation of the sample continued until failure.

b) Small Grain Size Sample at .120 in. Displacement Amplitude

At 1% of fatigue life, 200 cycles, irreversible grain boundary sliding was observed as well as grain boundary migration

similar to the case at the smaller amplitude after 1000 cycles. Accumulation of fatigue damage proceeded at a nearly similar pace as in the previous case in terms of percentage of total life. However, the damage did not spread to the neutral axis until after 8% of the life while at the smaller amplitude this had occurred by 4% of the expected life. This is shown by the two pictures in Fig. 4. The level of damage at 4% of the life for a small grain size sample cycled at a displacement amplitude of .060 in., however, was somewhat lower than the damage at 8% of life for a sample of the same grain size cycled at 0.120 in. amplitude. With the exception of these small differences of rate in the early stages of damage accumulation, the sequence of events observed for the 0.120 in. amplitude tests were also followed in the tests at 0.060 in. amplitude. Cracks at the surface were again not observed before 30% of the fatigue life.

c) Larger Grain Size Samples

Many of the features of fatigue damage accumulation in large grain samples for either a deflection amplitude of 0.120 in. or 0.060 in. were similar to the events which occurred in the smaller grained samples. However, certain qualitative differences were thought to be significant.

Grain boundary migration was not so extensive as that in small grain size samples. Instead of migrating over their entire length, grain boundaries which were initially nearly parallel or

perpendicular to the tensile axis tended to become serrated in shape with the legs of the serrations lining up near 45° to the stress axis. Frequent alignment of the slip striations which originate from the corners of serrations with the direction of boundary segments were observed. Extensive curvature of the initially flat grain surfaces during earlier stages of deformation indicated that plastic deformation inside individual grains may have occurred at an earlier stage of life. However, this may have been primarily because curvature was easier to detect in the larger grains. Large slip steps inside of grains were first observed at approximately the percentage of life where deformation inside of grains became noticeable in the small grain size samples. A typical example of the damage occurring on the large grained sample is shown in Fig. 5.

B) Metallographic Observation of Sectioned Specimens

Samples run for certain percentages of their expected fatigue life were sectioned, polished and etched to find if any cracks had appeared inside the material. Observations were made with an optical microscope and with a scanning electron microscope. Since the observations made on the surfaces of the samples showed no large differences between the events which occurred at different amplitudes, the metallographic observations were all made on samples cycled at .060 in. amplitude.

a) Small Grain Size Samples

No interesting features were observed in sections which were made from a sample cycled up to 8% of its expected fatigue life. No cracks were observed and no voids were detected.

After 20% of the expected life many interesting internal features were observed using the optical microscope.

Voids and cracks were observed along grain boundaries as shown in Fig. 6. It was confirmed that these cracks and voids resulted from the fatigue process because unstrained samples did not show these features. Many intergranular cracks were observed around the minimum section. In some cases these cracks were rather long, but in other cases they seemed to be nearly circular voids. Therefore, it may be assumed that the intergranular cracks originate as voids and grow as the voids link together. See Fig. 7.

At high magnification in the scanning electron microscope, many particles were observed along grain boundaries as shown in Fig. 8. In some cases these particles showed no evidence of separation from the matrix. However, around many particles cracking was observed as shown in Fig. 9. A typical size of the voids is 1 micron diameter.

Subgrain boundary formation was observed near the minimum section, but was confined to a smaller region than was the void formation and the intergranular cracking. A limited amount of triple point cracking was observed.

b) Large Grain Size Samples

In a previous study no significant features were observed on sections of large grained samples until they had been fatigued up to 20% of their expected fatigue life. Therefore, the only sections made in this work were on a sample fatigued for 20% of its life.

Intergranular cracking was observed as shown in Fig. 10. When viewed at a magnification of 20,000 x, individual holes were observed at the bottom of the continuous crack along the grain boundary at the etched surface. Less particles were observed along grain boundaries in the large grain size sample than in the small grain size sample. Therefore, the cracking which was observed was often not associated with a particle. Polygonization was more wide spread than in the small grain size samples (Fig. 13a). However, polygonization seemed to be confined to those regions where large amounts of plastic strain inside the grains would be required to accommodate the grain boundary sliding. For example, more subboundary formation was observed around crooked boundaries (Fig. 13b) than along the straight grain boundaries. Again limited triple point cracking was observed.

DISCUSSION

A) General Phenomenon

In observing the outside surfaces to see any surface crack initiation and subsequent inward propagation of the cracks, the difficulty is in finding an observation technique which can detect the cracks. The replication method followed by use of the scanning electron microscope to examine the replicas suffers from its inability to distinguish between a deep crack and a shallow notch. As shown in this work fatigue damage, especially out of plane grain boundary sliding creates many notches and folds which can easily be interpreted as cracks.

A stereo zoom optical microscope with large depth of field and magnification up to 200x does not provide as great a magnification or resolving power as the scanning microscope but greatly exceeds the replica method in versatility. A particular convenience was the ability to tilt the sample to any orientation and to change the lighting conditions to look down into grooves and corners of steps to determine their depth. By making observations with the optical microscope, it was possible to determine unambiguously that many crack-like features which were formed early in life were in fact not deep cracks and did not deepen during further cycling.

This improvement in observation technique showed that cracks did not initiate on the surface until the specimen was fatigued up to 32% of its expected fatigue life. Widespread secondary cracking away from the final failure which was reported to have been observed on

Silicon-Iron fatigued under similar conditions (24) were found to be relatively shallow notches rather than cracks. Therefore it can be concluded that crack initiation is not as easy a process relative to crack propagation as previously reported.

Metallographic observations of the sectioned specimens, which were run up to 20% of their expected fatigue lives showed intergranular cracking and cavitation, even though no cracks were observed at the surface by this stage of life. Therefore it appears that fatigue failure under the test conditions used in this study begins by intergranular cracking which starts from inside the specimen away from the free surface.

Among the other features of the behavior of the material as observed on the surface during the test, grain boundary migration and plastic deformation inside the grains seem most important. The creation of curvature and bulges on the originally flat surface as a result of the large plastic deformation inside the grain was observed more frequently in large grain size samples than in small grain size samples. On the other hand, grain boundary migration was found more often in small grain size samples than in large grain size samples. It would therefore appear that the incompatibility of grain boundary sliding is accommodated by plastic deformation in the large grain size samples while grain boundary migration to easy sliding orientations relieves much of the accommodation problem.

In all cases studied, the accumulation of fatigue damage proceeds in a manner which is nearly proportional to the percentage of fatigue life expended. In other words, samples which have been cycled to the same percentage of their total life show a similar accumulation of damage regardless of the displacement amplitude or the grain size. Thus the entire process of fatigue appears to be accelerated by nearly the same factor by an increase in the cycle amplitude. This is clearly true from the photographic records to within a factor of two on cycles, since a sample which has been run for 16% of its life at one strain amplitude shows clearly less damage than another sample run for 32% at another strain amplitude but more damage than one run for 8% at the other strain amplitude.

There was no effect of grain size or the strain amplitude on the general features of the fatigue damage observed on external surfaces.

B) Crack Initiation Mechanism

Among mechanisms which have been suggested for crack initiation, void formation by the diffusion and condensation of vacancies is the oldest and most widely accepted theory. From the observation that no voids are observed in creep under compression it is assumed that the generation of vacancies is aided by tensile stress. The diffusion of vacancies to voids can occur by either of two mechanisms, grain boundary diffusion or lattice diffusion. However, Hull & Rimmer have shown that even for very large voids of the order of 10^{-2} cm diameter, the contribution from lattice diffusion is only 6% of the total. This is

calculated from the equation

$$\frac{N_L}{N_g} = \frac{D_L \cdot \rho}{D_g \cdot \delta z} \qquad (1)$$

where N_L and N_g are number of atoms removed from the void by lattice diffusion and grain boundary diffusion, respectively. D_L and D_g are, respectively, lattice diffusion and grain boundary diffusion coefficients, ρ is the void radius and δz the grain boundary width. In view of this it is presumed that the atoms are released from within a ring round each void, which lies in and has the same thickness, δz, as the grain boundary. The rate of growth of a void is determined by the gradient of chemical potential, ∇f, in the plane of the grain boundary. Since the diffusion flux, j, is given by

$$j = \frac{D_g}{KT\Omega} \nabla f \qquad (2)$$

and the potential is given by

$$f = -\sigma_n \Omega \qquad (3)$$

where σ_n is the normal stress acting across the boundary. The stress is assumed to be σ well away from any void. On the surface of a void

$$f = -2\gamma\Omega/\rho \qquad (4)$$

where γ is the surface energy. Hence as a rough approximation

$$f \simeq -\frac{\Omega}{a}\left(\sigma - \frac{2\gamma}{\rho}\right) \qquad (5)$$

where a is void spacing.

This gives

$$j = \frac{D_g}{KTa}\left(\sigma - \frac{2\gamma}{\rho}\right) \qquad (6)$$

This shows that only those nuclei for which $\rho > 2\gamma/\sigma$ will grow. For voids smaller than this critical size the applied stress is not sufficient to overcome the surface tension and sintering occurs.

In this experiment tests were conducted at constant strain amplitude, not under constant stress amplitude. But an order of magnitude calculation can be carried out by taking $\gamma = 1500$ erg/cm^2 and taking $\sigma = 2\times10^3$ psi which is the yield stress at 900°C. This gives $r = 0.5$ micron. This would appear to be an excessively large nucleus to be formed by vacancy condensation alone without aid from the other nucleation mechanisms which are discussed later. Therefore these voids are similar to those observed in nimonic alloys. This nucleus size is considered to be in very good agreement with the observed void size of 1 micron. Furthermore some grain boundaries transverse to stress axis were observed to be heavily cavitated, as predicted by the vacancy diffusion mechanism.

Another mechanism of void formation which has been suggested by many investigators is nucleation by grain boundary sliding. Extremely high vacancy supersaturations would be necessary to give spontaneous homogeneous nucleation of voids, but two types of sites at grain boundaries may be suitable for heterogeneous nucleation. These sites are ledges and particles.

Nucleation at ledges has been estimated by McLean to be possible if ledges are at least four atoms high. Also this nucleation process may be expected to produce a fracture surface with linear arrays of elongated cavities having a similar regularity and spacing to the broad slip bands found on the surface. Nucleation at particles would lead to a random dispersion of voids which are essentially equiaxed.

The observations seemed to be more in agreement with nucleation at particles than at ledges. In small grain size samples, many particles were observed at grain boundaries and many cracks were clearly associated with inclusion particles. Some voids were considered to be too large to have been formed by vacancy condensation because of the large number of vacancies which would be required. However, these voids were observed at a late stage in life and could have been enlarged by plastic deformation. No evidence of a regular array of ledges was detected. However in large grain size samples few particles were observed on boundaries and very few were observed in the intergranular cracks. Therefore the initiation mechanism for voids and cracks could not be determined, but the effect of inclusions would appear to be less important than in the small grain size samples.

In both grain size samples, the maximum level of cavitation was observed on grain boundaries with orientations near the maximum shear stress directions. This strongly indicates that sliding along the boundaries is an important part of the process.

Some investigators have proposed the possibility of grain boundary oxidation as the cause of voids. This seems to be unacceptable

because no relation between the amount of cavitation and the distance from the outside surface was observed. In the author's opinion, if the void formation is the result of grain boundary oxidation the voids should prefer those grain boundaries near the surface rather than those grain boundaries away from the surface.

The possibility of Zener cracking at triple points as a crack initiation mechanism under these test conditions is doubtful simply because the cracks observed were not predominantly located at triple points.

It seems that no single mechanism can explain the crack initiation mechanism of Silicon-Iron at elevated temperature fatigue conditions. Nucleation by grain boundary sliding is supported by the observation that the amount of cavitation is maximum near maximum shear stress orientations. However a few seriously cavitated grain boundaries which were transverse to stress axis were observed which may indicate the importance of the tensile stress component as in the vacancy condensation model.

The author believes that grain boundary sliding is the preferred mechanism of void nucleation while both grain boundary sliding and the vacancy diffusion and condensation contribute to the process of void growth. Other factors which have not been discussed can affect the preference for one of the two mechanisms. Local properties such as the grain boundary structure and dislocation interactions with the grain boundary could also be important.

CONCLUSIONS

1. Crack initiation in Iron - 3% Silicon at 900°C is not an easy process relative to crack propagation. Therefore the fatigue mechanism in Silicon-Iron can not be said to be crack propagation controlled.

Cracks initiate inside the specimen by cavitation along grain boundaries linking together to form intergranular cracks.

2. Crack initiation due to grain boundary oxidation is doubtful since there is no correlation between the level of cavitation and the distance from the surface.

3. It is likely that voids are nucleated at particles or ledges due to grain boundary sliding.

4. The void growth mechanism is likely to be a combination of grain boundary sliding and a vacancy diffusion process.

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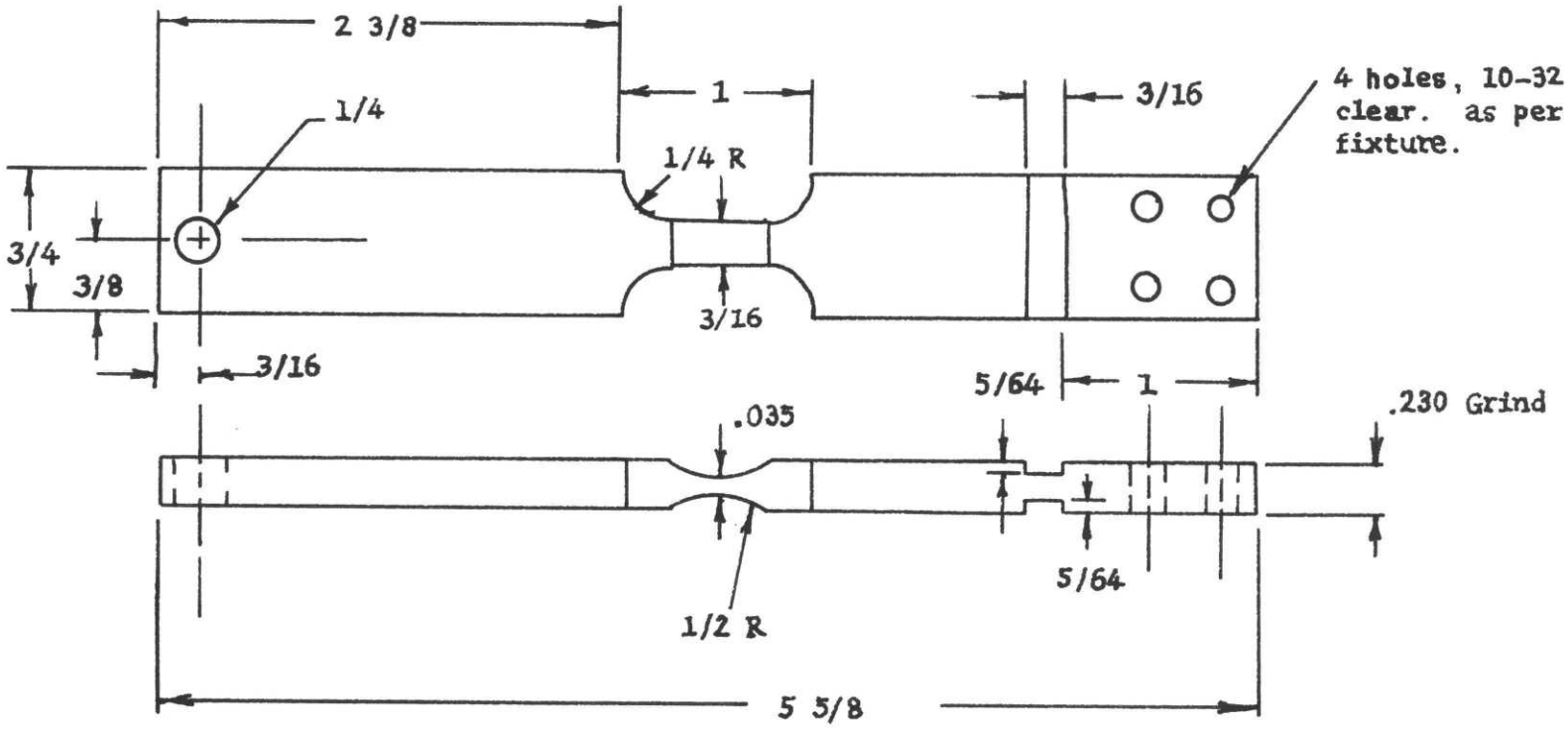


Figure 1 Specimen Design; Full Scale.

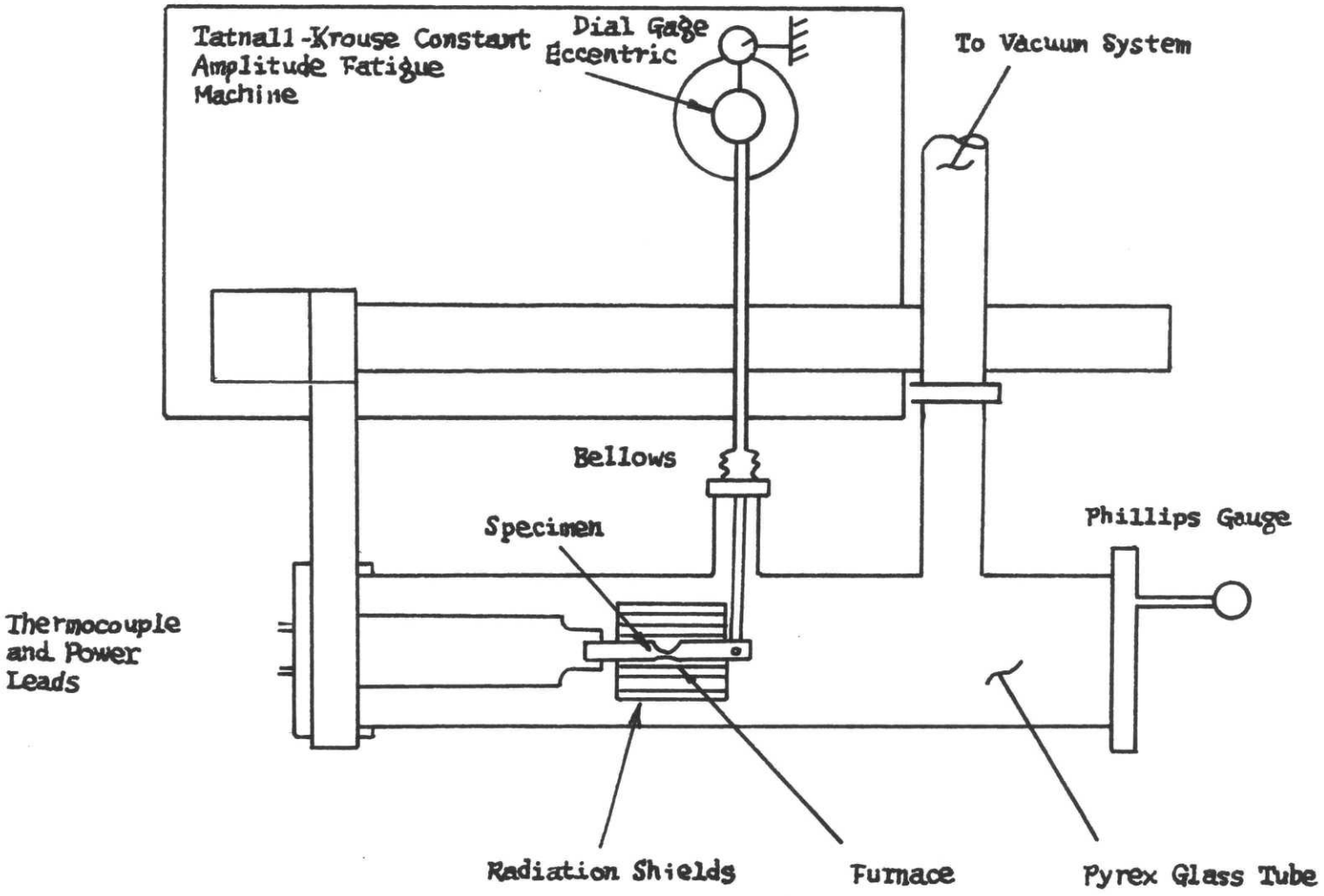
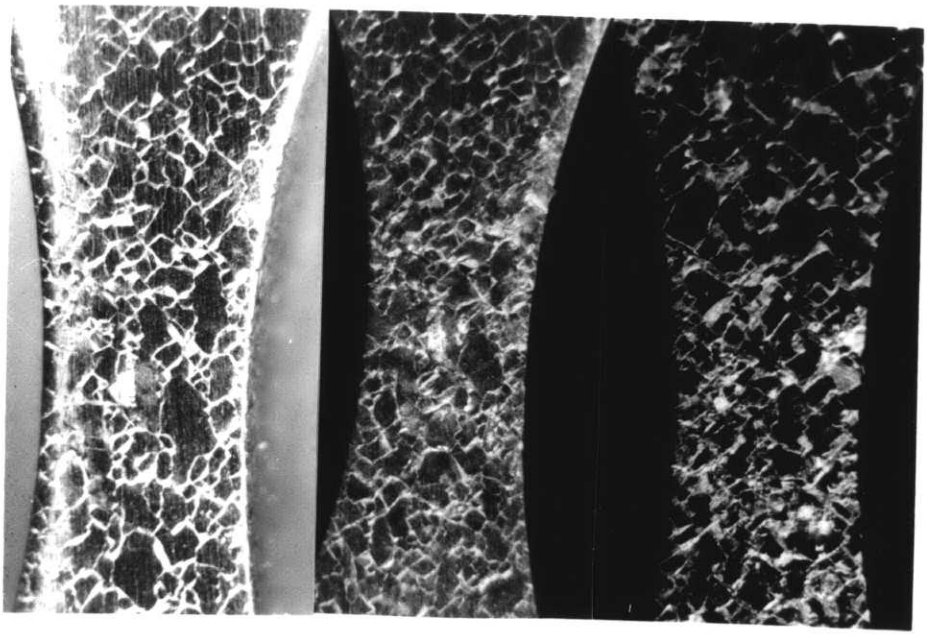
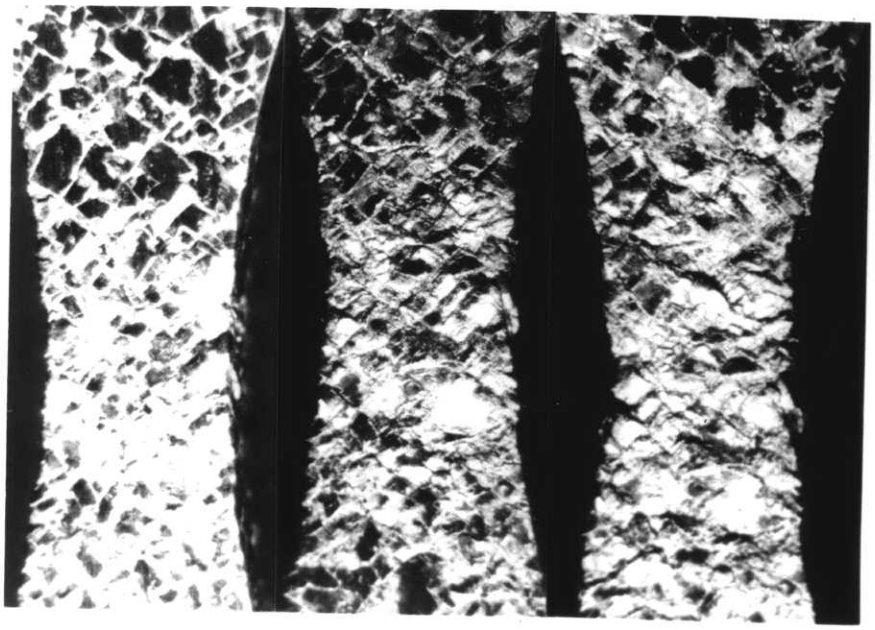


Figure 2. Schematic of Experimental Apparatus, No Scale

Fig. 3. Fatigue Damage of Small Grain Size Sample After Cycling to (a) 1%, (b) 2%, (c) 4%, (d) 8%, (e) 16% and (f) 32% of Total Life at .060 in. Displacement Amplitude.





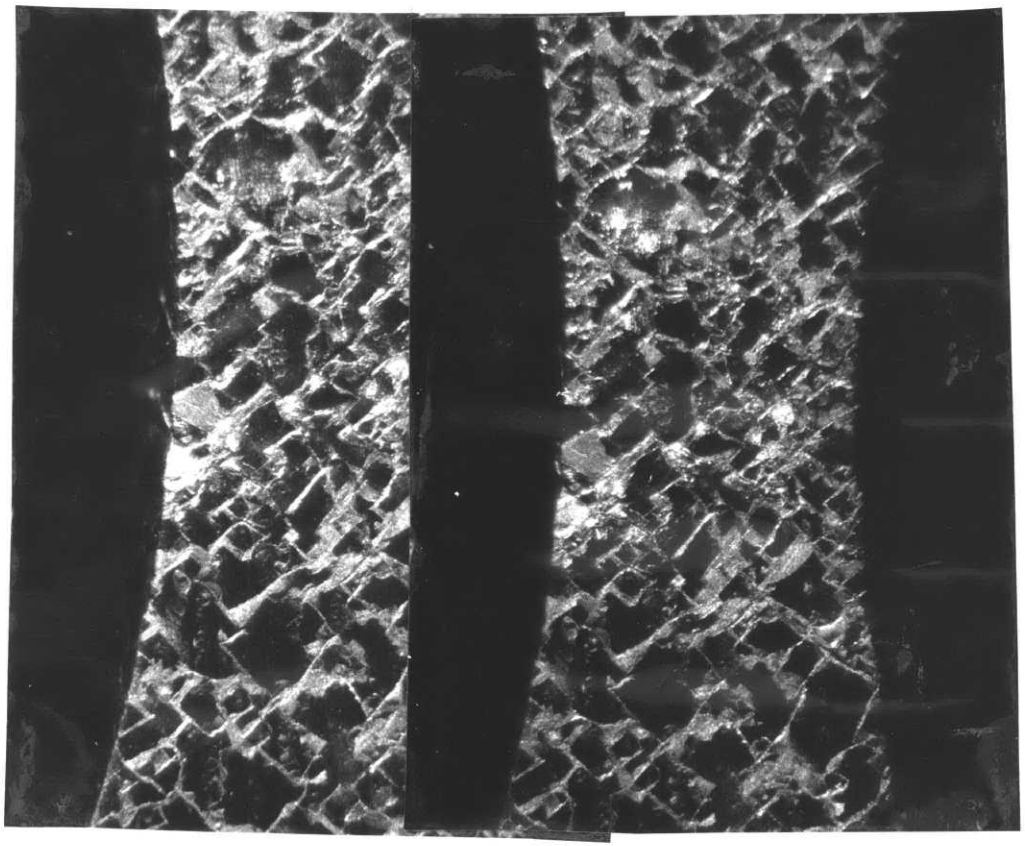


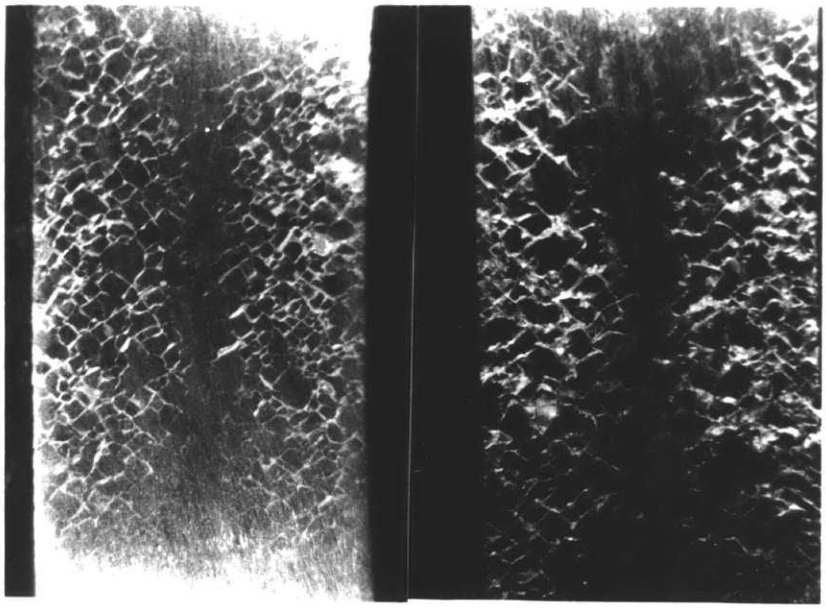
Fig. 4. Stereo Photograph of the Surface of Small Grain Size Sample Cycled at a Displacement Amplitude at .060 in. for 8000 Cycles (4% of fatigue life).

Fig. 5a. Side Surface of Small Grained Sample Cycled at .060 in. Displacement Amplitude for 4% of its Fatigued Life Showing the Beginning of Damage Along the Neutral Axis.

Fig. 5b. Side Surface of Sample Cycled at 0.120 in. Amplitude Showing no Significant Damage Near the Neutral Axis.

Fig. 6a. Fatigue Damage on the Top Surface of a Large Grain Size Sample Cycled for 5% of its Fatigue Life (5000 cycles) at .060 in. Amplitude.

Fig. 6b. The Side Surface of the Same Sample as Fig. 6a.



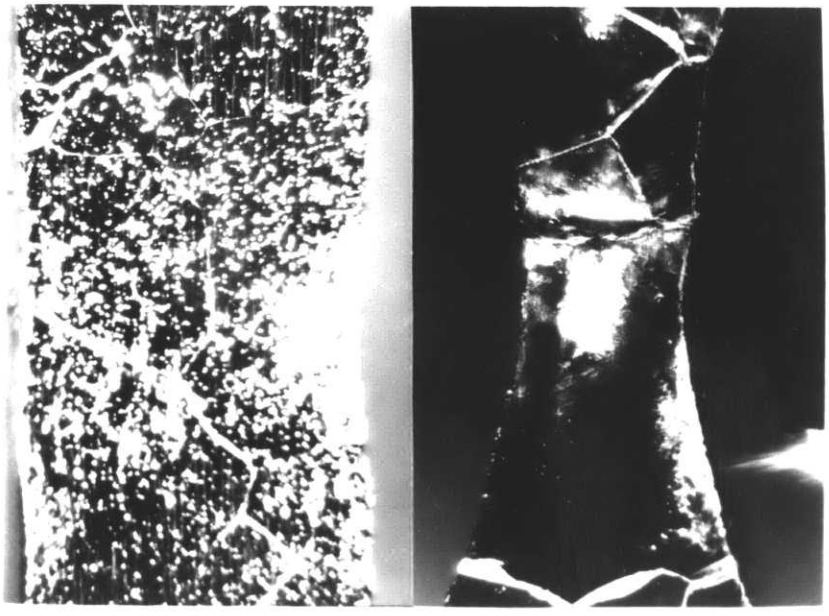


Fig. 7a. An Intergranular Crack Lying Perpendicular to the Tensile Axis on a Polished and Etched Section of a Sample Cycled 20,000 Cycles at .060 in. Amplitude (128x).

Fig. 7b. Interlinking of Voids to Form Intergranular Cracks as Revealed on a Polished and Etched Section of a Sample After 20,000 Cycles at .060 in. Amplitude (5000x).

Fig. 8. Uncracked Particles Along a Grain Boundary on a Polished and Etched Section (2000x).

Fig. 9. Cracks Around Particles on a Grain Boundary Revealed by Sectioning, Polishing, and Etching a Sample After 20,000 Cycles at .060 in. Amplitude (5000x).

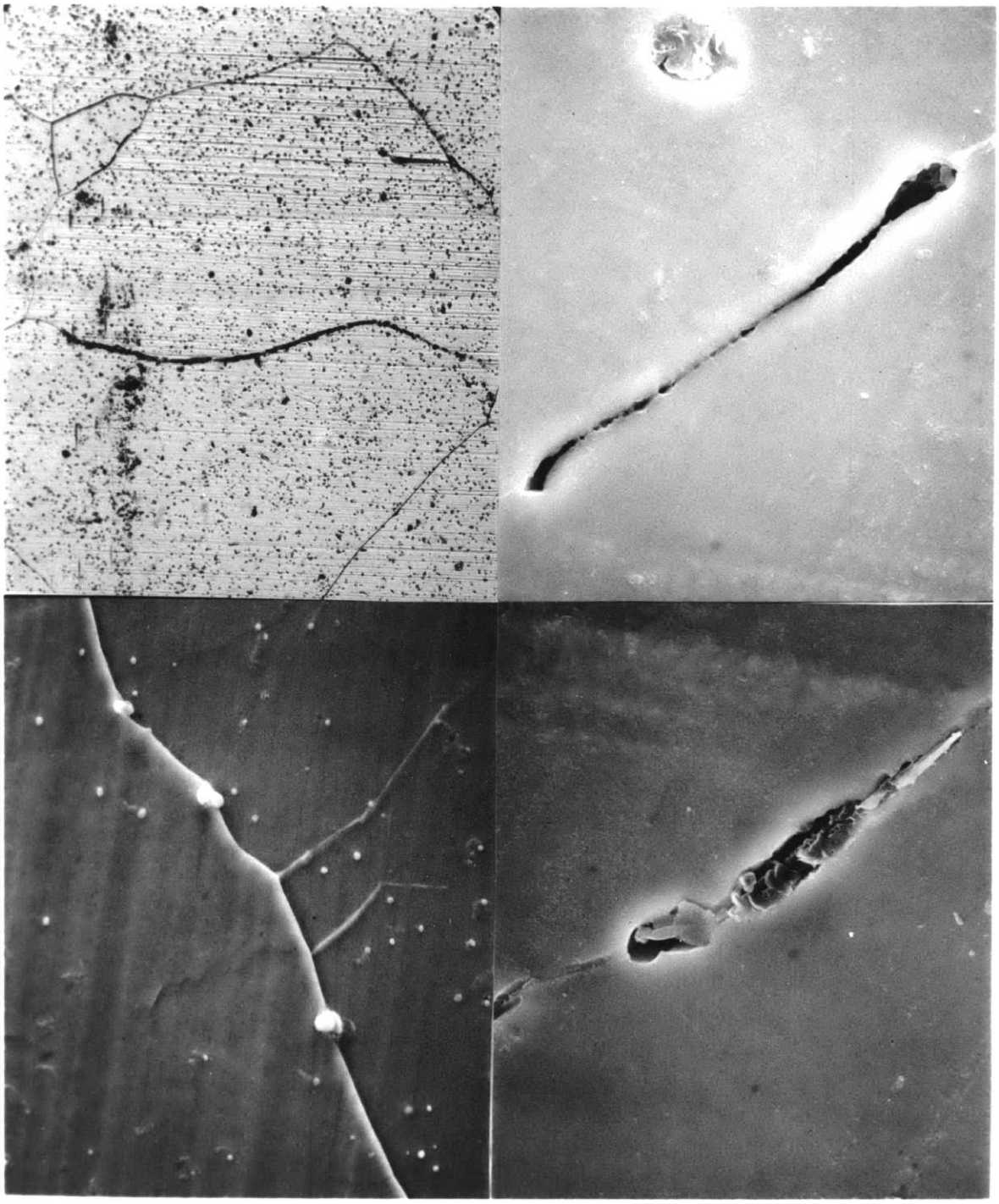


Fig. 10a. Intergranular Crack on a Large Grain Size Sample Fatigued for 20,000 Cycles at .060 in. Amplitude (2000x).

Fig. 10b. Holes Inside the Crack Shown in Fig. 10a.

Fig. 11a. Subgrain Formation Around a Surface Crack in a Large Grained Sample After 20% of the Fatigue Life (80x).

Fig. 11b. Subgrain Formation Near a Crooked Grain Boundary in the Same Sample as Fig. 11a (400x).