A STUDY OF THE MODE OF FATIGUE FRACTURE IN ANNEALED AND BURNISHED COPPER

Thesis for the Degree of M. S.

MICHIGAN STATE UNIVERSITY

V. Ramachandran

1962

This is to certify that the

thesis entitled

A STUDY OF THE MODE OF FATIGUE FRACTURE
IN ANNEALED AND BURNISHED COPPER

presented by

V. RAMACHANDRAN

has been accepted towards fulfillment
of the requirements for

M.S. degree in METALLURGICAL ENGINEERING

William E. Taylor
Major professor

Date August 17, 1962

O-169

ABSTRACT

A STUDY OF THE MODE OF FATIGUE FRACTURE
IN ANNEALED AND BURNISHED COPPER

by V. Ramachandran

The effect of varying the strain amplitude on the mode of torsional fatigue fracture has been studied in annealed and burnished copper. In annealed copper, low strain amplitudes cause a transgranular slip-dependent fracture whereas at high strain amplitudes, the fracture is intergranular. There exists an intermodal transition point at which both types of fracture are possible. In burnished copper, the fracture is intergranular at all strain levels investigated.

Burnishing enhances the fatigue life of copper at low strain levels, perhaps by inhibiting the formation of slip lines and associated cracks and also retarding the propagation of cracks. However, at high strain levels, the burnished metal fails earlier than the annealed metal. Further work is necessary to find the appropriate mechanism for this.

A STUDY OF THE MODE OF FATIGUE FRACTURE

IN ANNEALED AND BURNISHED COPPER

By

V. Ramachandran

A THESIS

Submitted to
Michigan State University
in partial fulfillment of the requirements
for the degree of

MASTER OF SCIENCE

Department of Metallurgy, Mechanics and Materials Science

1962

ACKNOWLEDGEMENT

The author is very much indebted to Dr. W. E. Taylor for his guidance and many thought-provoking discussions in connection with this work. He is also grateful to Dr. C. A. Tatro for his very valuable assistance in instrumentation. Mr. B. Curtis and Mr. D. Childs of the Division of Engineering Research are thanked for their help in machine shop work. The author wishes to acknowledge the help of Dr. A. J. Smith, Professor and Head of the Department of Metallurgy, Mechanics & Materials Science for affording all the facilities necessary for the completion of this project and finally he wishes to express his gratitude to the United States Agency for International Development and the Government of India for offering a scholarship for studies and research at the Michigan State University.

TABLE OF CONTENTS

LIST OF FIGURES

I. INTRODUCTION

Metal fatigue has been a very intriguing problem to design engineers and research workers and is still one of the most challenging fields for engineering progress (1). Fatigue in metals is referred to as a process of progressive fracture. Under repeated stressing, a steady propagation of damage takes place, crystal by crystal, throughout the metal member until a large portion of the critical cross section has its structure sufficiently disrupted to initiate microscopic cracks. These microcracks, on continued cycling, grow, join and form a visible crack leading to failure.

Several aspects of fatigue have been investigated in great detail and excellent reviews on the subject are available today (2, 3, 4). The investigations were concentrated on the macroscopic, microscopic and the atomistic levels and the old theory of fatigue failure by the mechanism of "crystallization" and embrittlement of metal had to be modified in the light of the results of these recent investigations (5).

The well known S-N diagrams are a result of the investigations dealing with the phenomenological behavior of metals on the macroscopic level where the stress and the corresponding number of cycles to cause fracture are the main concern and form the practical basis for design work. The microscopic and submicroscopic phenomena in metals during fatigue loading have been studied by various workers, resulting in a fund of useful information leading to an understanding of the mechanism of fatigue fracture. The third and the last approach was on an atomistic level, aimed at the explanation of fatigue on the basis of solid state theory, dealing with lattice defects. Among the defects in solids, dislocations play a vital role in the deformation and fracture of metals and the dislocation theory aims at an explanation of fatigue of metals also.

In recent research, attempts have been made to obtain a good correlation between the microscopic and submicroscopic observations of metals during cyclic stressing and their macroscopic behavior. The macroscopic or the phenomenological behavior of metals in fatigue, as mentioned earlier, are best studied by constructing the S-N diagram from experimental data. From this diagram, the fatigue life or the number of cycles of failure at a particular stress level can be determined. Again, for a particular number of cycles of service, the maximum stress that can be safely applied without failure can also be computed from the same diagram. Materials like steel exhibit a definite "endurance limit" which is the limiting stress below which the material can endure an indefinitely large number of cycles without failure. Many nonferrous metals do not have a definite endurance limit. However there is a knee in the S-N curve and a pseudo-endurance limit may be found for such materials.

A closer examination of the S-N curve indicates that as the number of cycles increases, the stress to failure drops rapidly at first and then gradually later. After crossing the knee of the curve, the rate of drop is very small. A hitherto unsuspected discontinuity in the S-N curve has been recently reported by Porter and Levy (6). Metallographic observations have confirmed that the discontinuity corresponds to a change in the mode of fatigue failure. It was believed that the mechanism of deformation and failure changes as the amplitude of cyclic stress increases, the general shape of the S-N curve arising from the superposition of two distinct mechanisms. If the fundamental mechanism for fatigue is different in the two regions, the path of the fatigue crack also may be different. It is often convenient to study the fatigue process at constant strain amplitude rather than constant stress and a study of the mode of fatigue fracture at high and low strain amplitudes in annealed copper forms a part of the present investigations.

Surface cold working plays a prominent part in the fatigue behavior of metals. There are many instances where very beneficial effects have been obtained by surface cold working, mainly an enhancement of fatigue life. Surface cold working introduces residual compressive stresses on the surface layers and the enhancement of fatigue life is generally attributed to the presence of these compressive stresses. Recent metallographic investigations by Warke and Taylor (7) have indicated that surface cold working by shot peening inhibits slip band formation in the surface layers and also retards the propagation of microcracks which are formed very early in the life of a fatigue specimen. In annealed metal, at low cyclic stresses, a fatigue crack originates at a fissure formed in a slip band and generally follows the slip band pattern. Since the surface cold working operation inhibits slip band formation, the mechanism of formation and propagation of fatigue crack in surface cold worked metal may be different from that for annealed metal. A study of this aspect is the major objective of the present work.

Industrially, surface cold working is achieved by various processes such as shot peening, surface rolling and burnishing (8). Burnishing leaves the metal surface in a smooth condition enabling easy metallographic work and hence burnishing was chosen as the method for surface cold working the metal and the study was extended to high strain levels also.

The next section describes in detail the mechanisms operating in the high stress and the low stress fatigue. Chapter III deals with the experimental technique of the present investigations and is followed by a detailed discussion of the results and the possible mechanisms.

II. BACKGROUND AND THEORY

In a discussion of the mechanism of fatigue at various stress or strain levels, it is desirable to consider the simplest form of fatigue. In the following discussion, simple fatigue of pure metals will be considered. The cycles of strain will be symmetrical about zero mean strain. The metals will be initially annealed and hence free from superimposed internal strains.

Generally, the fatigue crack is initiated near the surface of the metal (9). Ewing and Humphrey were the first to study the microstructural changes on the surface of metals during cyclic stressing (10). They suggested an "attrition" theory of fatigue in which the repeated application of an unsafe stress produces repeated slip, resulting in attrition of the metal along the slip planes, eventually leading to cracking and failure. Gough (11) subjected various metal single crystals to cyclic stressing and his metallographic observations indicated that fatigue is a process of strain hardening by slip on the operative glide planes leading to cracking on these planes. Gough also suggested that slip is accompanied by fragmentation of the original crystal into small crystallites in the vicinity of the active slip planes. According to Orowan (12) also, work hardening takes place on the slip planes and cracking occurs when the stress in the strain hardened region exceeds the rupture strength of the metal.

According to the present day knowledge, the fatigue process can be divided into three distinct stages, based upon experimental evidence. These stages are well manifested in tests employing a constant strain amplitude. In the first stage of fatigue of annealed metals, slip takes place in the active slip planes, leading to work hardening of the metal to a hardness level which depends upon the metal and the strain amplitude. On the other hand, if a cold worked metal is subjected to cyclic stressing, the hardness decreases in the first stage (13, 14, 15). The rate of work hardening in cyclic straining is found to be less than that in unidirectional straining. In the work hardening range, slip lines are formed throughout the metal, dispersed uniformly. With increase in strain amplitude, more slip lines are formed in a larger number of grains. If the metal is reannealed at this stage, the metal is brought back to its original annealed condition and this method has been suggested as a means of enhancing the fatigue life of originally annealed metals (16). Depending upon the strain amplitude, work hardening saturates and stops sooner or later during the early part of fatigue life. At high strain amplitudes, fracture may precede the saturation hardening. Work hardening of annealed metal in the first stage has also been proved by x-ray diffraction (17, 18).

The second stage starts after saturation hardening and extends over the greater part of the life. In the second stage, there is no further hardening, but plastic deformation continues. No change in the x-ray diffraction pattern takes place. Few new bands form, the majority of which being confined to planes adjacent to the original slip lines. Thus the originally formed slip bands become wider. The metal in the slip bands, due to alternating strain, eventually is raised or lowered forming grooves and ridges, also referred to as intrusions and extrusions (19). The formation of intrusions and extrusions has been established beyond doubt by recent electron microscopic work (15, 20, 21). Thompson (22) electropolished the specimens at this stage and after removing a few microns thickness of metal, found that most of the slip bands are removed by the polishing operation while some bands still remain and these are referred to as persistent slip bands. It is in these persistent slip bands that fatigue cracks originate.

Wood devised an ingenious method for studying the slip band topography of surface and subsurface layers (23, 24). He ground and polished a narrow longitudinal flat of width about .01" on the cylindrical surface of a fatigue specimen and obtained a mechanical magnification of the subsurface layers by about twenty times. This coupled with optical and electron microscopy produced very high magnifications (21, 25). Such studies have definitely established the formation of extrusions and intrusions, the origin for fissures. These fissures on continued cycling, form visible cracks which then propagate and cause fracture. It has been found that higher the stress amplitudes, the shorter and less numerous are the bands and at very high amplitudes, bands are practically absent. This suggests that at high stress levels, a different mechanism operates for fatigue failure.

The third stage of fatigue involves the propagation of cracks. The submicroscopic cracks coalesce to form visible spreading cracks resulting in ultimate fracture. This third stage in annealed metals is generally of short duration in comparison with the total number of cycles to failure. At low strain amplitudes, the path of the crack is transcrystalline, following the slip band pattern. At high strain levels, the bands are very few or almost absent. The crack originates in a grain boundary and generally follows an intergranular path. Even if it crosses a grain, it does not necessarily follow the slip band pattern.

Wood has shown conclusively that two essentially different types of failure can happen in fatigue, one predominating at high stress amplitudes and the other at low stress levels where the S-N curve tends to become parallel to the cycle axis (24). The high and low levels of stress are termed by Wood as the H and F ranges. The mode of deformation in the high stress or the H range is essentially the same as in static deformation. The deformation process is one of coarse slip (26) and appears to occur primarily through the operation of Frank-Read sources (27). Fracture in the H range is actually a delayed static fracture, delayed because of the hardening that precedes the crack nucleation.

In the low stress or the F range, the deformation is due to fine slip bands that appear faintly at first and then grow in strength gradually. Failure appears to be due to a gradual deterioration of the structure in the persistent slip bands where fissures start. Large amplitudes excite coarse slip or avalanche of slip movements through hundreds of lattice spacings. Low stress amplitudes excite fine slip, probably by the to-and-fro motion of a large number of individual dislocations rather than the avalanche burst of dislocations from Frank-Read sources which are responsible for coarse slip. The total plastic strain in the F range is abnormally high and this produces, along limited lengths of the slip planes, zones of abnormally high lattice distortion and these zones at an early stage of specimen life, not more than 1/10, turn into visible fissures. There is no significant hardening so the lattice distortion must be due to concentration of point defects such as vacant lattice sites or interstitial ions.

Limited metallographic evidence is available in the literature to confirm the two different mechanisms of fatigue at high and low stress amplitudes.

Kemsley studied the deformation markings during cyclic stressing of copper and observed numerous distinct striations at low stress amplitudes. These striations are identified with slip traces. At high stresses, striations were very few. He also concluded that localized deformation characteristic of fatigue loading occurs only at low stress amplitudes and at high stress levels, the deformation is less localized and exhibits features similar to those produced by unidirectional or static loading (28).

Both intercrystalline and transcrystalline fatigue cracks have been observed in copper specimens by Strom (29). Kemsley found that cracks are transcrystalline in low stress ($\pm$15,000 psi.) specimens and intercrystalline in high stress ($\pm$25,000 psi.) specimens (30). Forsyth studied the effect of stress amplitude on the deformation bands during cyclic loading of aluminum-1/2% silver alloy and found that at high stresses deformation bands of the kink type are formed and at low stresses, these kink type bands are absent but slip is predominant (31). Porter and Levy also observed intercrystalline fracture in copper on high stress ($\pm$25,000 psi.) fatigue and transcrystalline fracture at low stress ($\pm$15,000 psi.) levels. They were the first to investigate how the different mechanisms of fatigue are reflected in the S-N curve (6). Recent electron microscopic investigations on the fatigue process in copper by Bendler and Wood also confirm the existence of the two different mechanisms of fatigue at low and high stress amplitudes (21).

A unified engineering theory of high stress level fatigue has been proposed recently (32). This theory based on the dislocation theory of metals and the macroscopic elasto-plastic fracture theory predicts the number of cycles to failure at any stress level. This theory is, however, applicable only to the high stress range of the S-N curve, where the mode of deformation is definitely different from the low stress mode.

Due to very limited evidence in literature for the two types of mechanisms of fatigue, the present work was started to study the effect of constant strain amplitude on the mode of fatigue cracks in annealed and burnished copper and correlate the same with the ϵ-N curve.

Surface cold working processes for improving the fatigue life of metals are well known (8, 33, 34). Surface rolling is applied to railroad axles and shafts (35). Shot peening is used to improve the fatigue life of springs. Burnishing is employed in the automobile industry for wheel spindles. Among these methods, burnishing is the simplest operation needing very simple equipment. The operation leaves the metal surface in a smooth condition.

In burnishing, a hard steel roll is kept pressed against the metal surface under constant load. The burnishing tool can be fixed to the tool post of a lathe and passed on the entire length of the rotating metal piece. The metal is thus compressed plastically. As the surface layers are continuous with the metal underneath, residual compressive stresses are produced in the surface layers.

Until recently, it was believed that the enhancement of fatigue life by surface cold working is due only to the presence of residual compressive stresses in the surface layers, which tend to oppose the service stresses. Warke and Taylor (7) made metallographic studies on taper sections of shot peened copper bars subjected to alternating torsion of constant low amplitude. They have established that in the surface layers of shot peened bars, slip lines are scarcely formed and that the fatigue fracture even at very low strains is intergranular and not slip-dependent. The absence of slip bands was believed due to the prevention of dislocation movement in surface layers by the already existing complex dislocation network introduced by the peening operation. A similar mechanism may be expected to operate on burnished bars also. No data are as yet available on the effect of varying strain amplitude on the fatigue life as well as the mode of fatigue fracture in burnished metal and hence this aspect also has been investigated in the present work.

III. MATERIALS AND METHODS

The material used in the present studies was oxygen free high conductivity copper of 99.98% purity. The metal was obtained as 1/2" round cold drawn rods having a hardness of 92 Rockwell F. Fatigue specimens of the R. R. Moore design were machined from these rods, the diameter at the center being 0.300 inch. The gage length of the specimens was polished with emery paper varying in fineness from No. 0 down to No. 000. Care was taken to see that the surface was free from circumferential scratches other than those produced by the last abrasive.

The fatigue test specimens were annealed in a "Burrell" resistor heated tube furnace continuously flushed and filled with argon gas. The specimen was kept in a boat connected to a rod. The cold end of the furnace tube was closed with a one-holed stopper through which the rod was introduced. The temperature of the furnace was regulated by a Variac and maintained at 1250° F. After flushing the furnace tube with argon gas for five minutes, the boat containing the specimen was introduced into the hot zone of the furnace by means of the rod. The temperature of the furnace was measured by a thermocouple placed immediately above the specimen. After holding the specimen at 1250^{o}F for fifteen minutes, the boat was quickly withdrawn from the hot zone and the specimen was quenched in cold water. Thus all the specimens were annealed. The hardness of the specimens after this treatment was 50 Rockwell F.

After annealing, the specimens were divided into two batches. The specimens in one batch were electropolished using a "Buehler" Electropolisher. A two to one solution of methyl alcohol and concentrated nitric acid was used as the electrolyte. A hollow cylinder of stainless steel of 3/4" i.d. was used as the cathode for producing uniform polish all around the specimen. A bright, scratch-free, polished surface was produced in five minutes with a current of 2 amperes. This operation also etched the specimens just enough to reveal the grains of copper under a microscope. The samples were covered in fine tissue and stored in a desiccator to prevent undue tarnishing.

The second batch of the specimens were burnished uniformly. A special spring-loaded burnishing tool was made and all the specimens of this batch were burnished uniformly

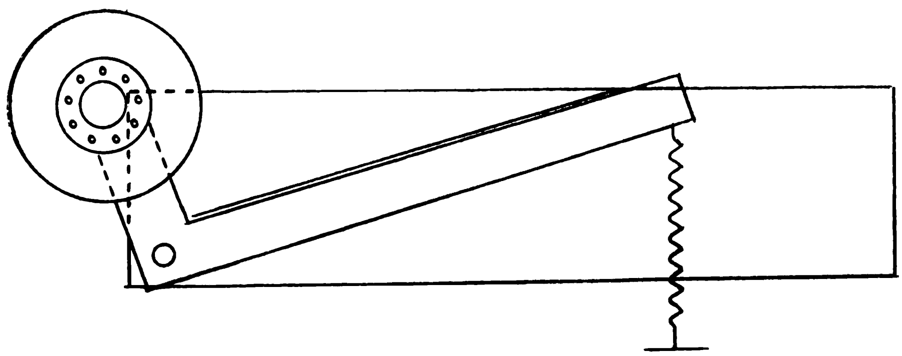

Figure 1. Constant load burnishing tool.

with this tool. The burnishing tool, shown in figure 1, was a hardened steel roller mounted on bearings. The roller-bearing assembly was fixed to one end of a lever which was pivoted to a tool holder block. The tool holder block can be readily fixed to the tool post of a lathe. The other end of the lever was connected to a spring. Using a spring gauge, a force of forty pounds was exerted on the spring which in turn moved the lever arm to a particular position. The position of the edge of the lever arm was scribed on the tool holder surface for reference. During actual burnishing operation, the other end of the spring was fixed to the base of the tool post. The fatigue specimen was mounted on the lathe and kept rotating. The burnishing roller was applied to the surface of the specimen. The tool post was moved following the same contour as the fatigue specimen, guided by a template. The entire gage length of the specimen was thus burnished. Throughout the burnishing operation, the lever arm was always kept with its edge coinciding with the scribed mark, thus insuring uniform pressure throughout each specimen.

The burnished samples were then electropolished in a similar manner and stored in another desiccator.

A constant strain amplitude torsional fatigue machine, shown in figures 2 and 3 was used to test the specimens. Two specimen holders were machined to match with the taper at the ends of the fatigue specimens. These shafts were mounted in an arbor. The rear shaft was held by a clamp and slid into the arbor. The front shaft was mounted on bearings driven by a lever arm. The other end of the lever arm was connected through ball bearings to an eccentric cam which was coupled to the shaft of a General Electric, 1/4 horsepower, 3450 rpm motor by means of set screws. The number of cycles of strain applied to the specimen were measured by a counter coupled to the rear end of the motor shaft.

For mounting the specimen, the rear shaft was first removed, one end of the specimen was introduced into the hollow front shaft and then the rear shaft was brought in to accommodate the other end of the specimen. The required cyclic torque was imparted to the specimen by the eccentric cam connected to the lever arm and the motor. A set of cams of different eccentricities was employed to impart varying torques to the specimens.

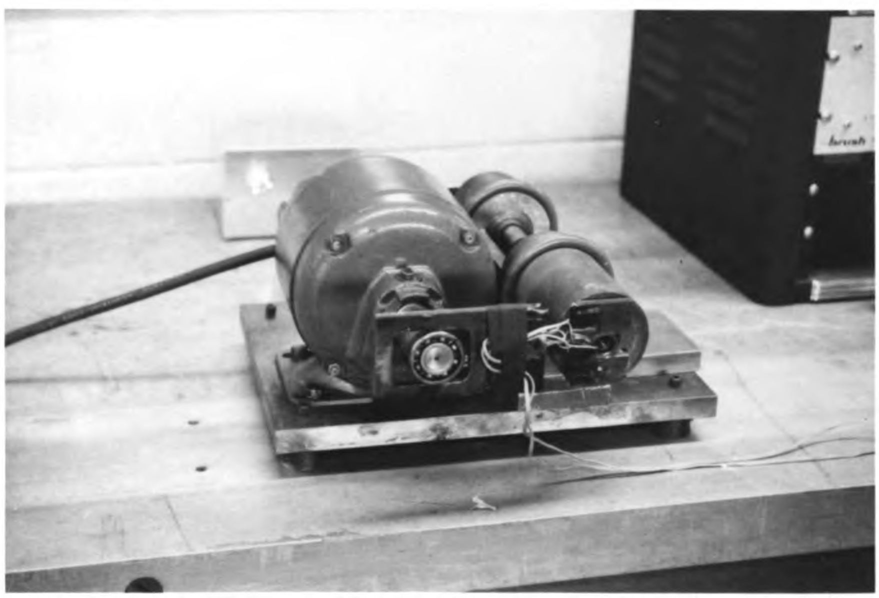

Figure 2. Constant strain amplitude torsional fatigue machine, front view.

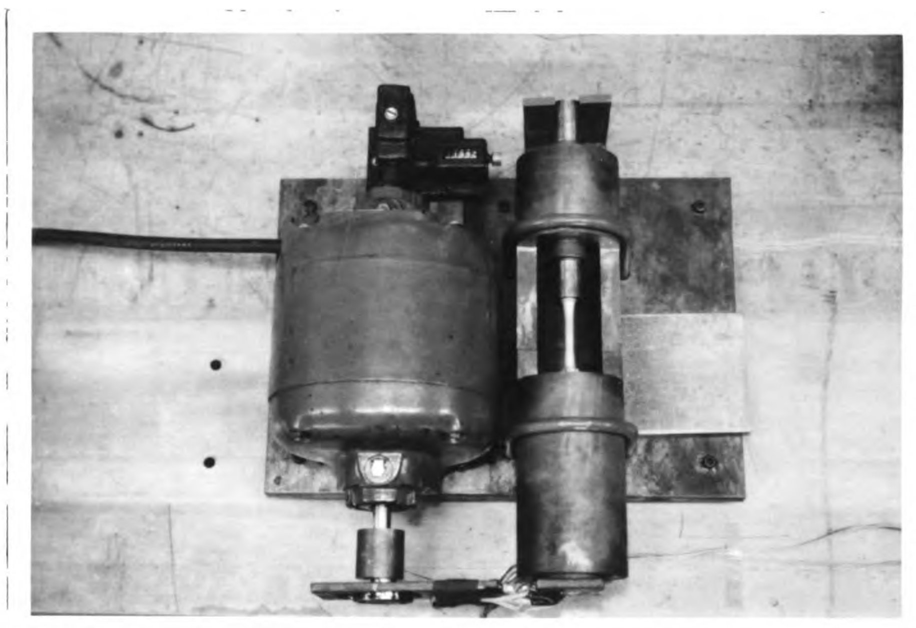

Figure 3. Constant strain amplitude torsional fatigue
machine, top view.

A relay system was employed to shut off the motor at the instant cracks were formed in the specimen. The fact that the torque on the specimen changes when a crack develops in it was made use of in operating the relay. Strain gauges were mounted on the constriction of the lever arm. The output from the strain gauges was amplified by a "Brush" Bridge Amplifier and then displayed on a "Tektronix" Model 104 RM 32 Cathode Ray Oscilloscope. The automatic shut off relay system was constructed as per the circuit diagram shown in figure 4 and the relay system was connected between the amplifier and the motor as indicated in figure 5. The resistance of the strain gauges after the introduction of the applied torque was balanced using the amplifier-oscilloscope circuit. The relay system shut off the motor as soon as cracks were developed in the fatigue specimen and the balance of the bridge thus disturbed. Then the motor was driven by throwing a manual switch till the specimen broke into two.

Both the annealed and the burnished specimens were subjected to fatigue testing at constant strain amplitude, in the shear strain range of 0.0035 to 0.0070 in./in.

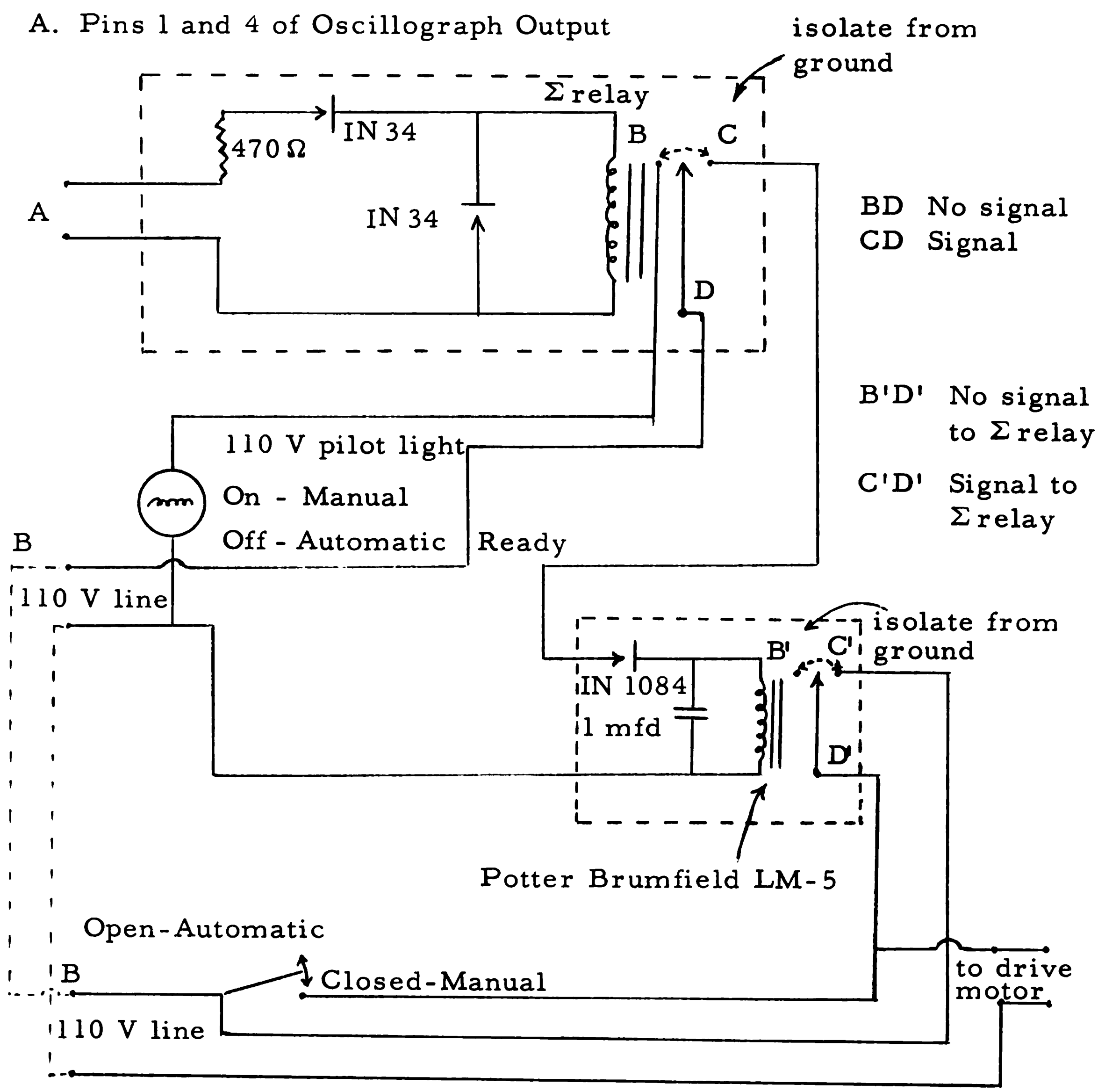

Figure 4. Circuit diagram for the automatic shut-off relay system.

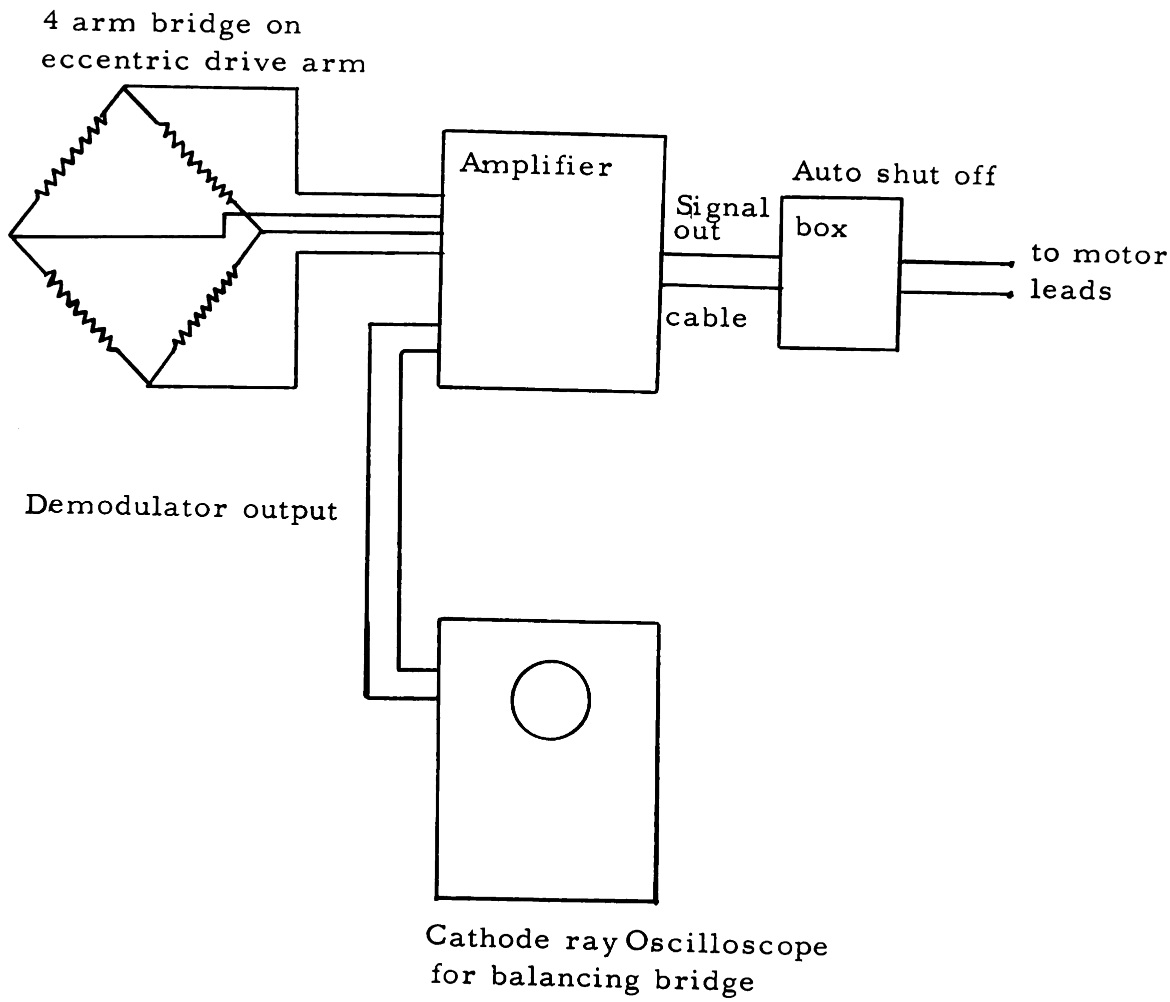

Figure 5. Assembly of the automatic shut-off relay system with the bridge circuit.

After fracture, the broken specimen was removed from the machine and the cracks on the surface were examined on a Bausch & Lomb Research Metallograph and photomicrographs taken with the same. The number of cycles to failure were noted on the counter.

IV. RESULTS AND DISCUSSION

A. FATIGUE OF ANNEALED COPPER

The effect of strain amplitude on the mode of fatigue fracture in annealed copper is illustrated in figures 6 through 10, At the highest strain, namely 0.007 in./in., the fracture is entirely intergranular. At the lowest strain of 0.0035 in./in., the fracture is entirely transgranular. As the strain amplitude is varied from 0.007 to 0.0035 in./in., there is a transition from intergranular to transgranular fracture. Thus two different modes of fatigue fracture exist, one at the strain amplitude of 0.007 in./in. and the other at 0.0035 in./in. These may be designated as the high-strain and the low-strain modes respectively.

I. Low Strain Fatigue

As a specimen is subjected to cyclic straining at low strain levels, slip lines first form and then develop into bands, described by Thompson et al. as "Persistent Slip" (36). Fatigue cracks are initiated in these persistent slip bands as shown by the work of Hempel (20), Porter and Levy (6), Warke and Taylor (7) and Bendler and Wood (21).

There is a local severe distortion along active slip planes which form the origin of fissures. Electron microscopic studies on the fatigue process in OFHC copper have proved that what appears under the optical microscope as continuous fissures are, actually, links of smaller discontinuous fissures, suggesting that the fissures responsible for final fatigue fracture originate in even more localized areas of abnormal distortion than the optical microscope would indicate. At low strain levels, the fissures initiated in the persistent slip bands grow following the path of the secondary slip bands which are formed in plenty at low strain levels. Hence the fatigue fracture at low strain levels may be expected to follow a transgranular path, following in general the slip band pattern. This is evident from figure 10, corresponding to a strain amplitude of 0.004 in./in.

II. High Strain Fatigue

At high strain amplitudes, the mechanism of crack formation and propagation appears to be different. According to Porter and Levy (6), initiation of crack occurs before the onset of secondary slip by the formation of a characteristic L-shaped nucleus, one limb of the L coinciding with a slip line and the other with a grain boundary. Also Z-shaped nuclei which are a combination of one slip line and two grain boundaries, can be the origin of cracks. The propagation of the cracks takes place at random, mostly following an intergranular path and even when crossing a grain, not being associated with slip lines. Porter and Levy consider Stroh's dislocation wedge model to explain the initiation of a crack at the end of a slip line (37).

Kemsley (28) observed deformation markings or striations in low oxygen copper subjected to low stress fatigue. The striations were found to be in abundance in low stress specimens ($\pm$11,000 psi) tested both in push-pull and as rotating cantilever. In high stress specimens ($\pm$20,000 psi), the striation density was almost zero. It was established that these striations are traces of slip packets. The absence of striations at high stresses may be due to a different mode of deformation from that occurring at low stresses. It was also established that a copper specimen deformed in tension to the same hardness as a fractured fatigue specimen, has a rumpled surface and exhibits no striations (38). This suggests that the mode of deformation is the same both in unidirectional loading and in high stress cyclic loading.

Louat has proposed a mechanism to explain the difference in the number of striations at low and high stress amplitudes (39). At low stress levels, in the first half cycle of loading, slip lines form with a spacing characteristic of the metal and the stress level; the greater the stress, the smaller the slip line spacing. During this slip line formation, dislocations generated by a source develop jogs, some of which are multiple, as a result of successive interactions with other dislocations crossing the slip plane. According to Louat, the drag caused by these jogs on the screw dislocations in which they form, leads to the formation of parallel lines of edge dislocations of opposite sign. The mutual annihilation of these opposite dislocations would be difficult if the jogs are multiple or if the result of the annihilation is the production of interstitials. The parallel edge dislocations act as obstacles around which other dislocations wrap themselves. Due to the stress field produced by this double pile-up around the original pair, some dislocations may be pushed past their twins and they may meet further dislocations. Some mutual annihilation may also occur and the resultant interstitials or vacancies are collected by dislocations in and around the central core of the pile-up, causing them to climb, resulting in a very complex dislocation network.

Due to the stress field around the double pile-ups, there will be a tendency for slip to occur close to the original slip plane. At distances away from the original slip plane, the stresses oppose the motion of dislocations and are inversely proportional to the distance from the slip plane. During unloading and reloading in the second half cycle, dislocation motion will be negligible. Repeated cycling would lead to increasing numbers of slip packets forming from single slip lines. At low stresses, the fatigue life is long. The cycles are repeated a large number of times to cause fracture. Hence there is plenty of opportunity for the formation of slip packets before fracture. If the stress amplitude is very high, the fatigue life is short and the development of slip packets takes place to a negligible extent. The same is the case in unidirectional loading, such as in tension. This explains the predominantly transgranular, slip-dependent fracture at low stress amplitudes and the absence of such fracture at high stress amplitudes.

Kemsley (30) has found evidence for the formation of valleys at grain boundaries. The marked surface rumpling in high stress specimens leads to the formation of valleys at the grain boundaries. These valleys can act as stress risers and lead to intergranular failure.

The mechanism of the low strain and high strain modes of fatigue fracture may be explained in a simpler manner as follows: It may be assumed that at low stress or strain amplitudes, the range of forward and backward movement of dislocations does not exceed the average spacing of lattice obstacles. Hence piling up of dislocations against these obstacles and the consequent strain hardening would be absent or very limited. These conditions favor continued slip movements in the region where they originally formed. Slip movements thus concentrate into bands producing local distortion in slip planes and also leading to fissures which then develop into cracks following the slip line geometry. At high strain amplitudes, the range of dislocation motion would exceed the average spacing of lattice obstacles. Hence, dislocations can pile up against these obstacles resulting in strain hardening. This, in turn, would disperse dislocation movement and also prevent the slip concentrations and localized distortion which is characteristic of low strain amplitude fatigue. Bendler and Wood (21) have experimentally shown that in high twist amplitude fatigue of copper, the fracture starts from microcracks generated by stress concentration at the dislocation pile-ups.

III. The ϵ-N Curve

In fatigue tests employing constant strain amplitudes, the results are best plotted as the ϵ-N curve, where ϵ is the strain amplitude and N the number of cycles to failure. Figure 11 is such a curve for annealed copper. Porter and Levy (6) have shown that the usual S-N curve for annealed specimens can be modified and drawn as two distinct curves, one for stresses above and the other for stresses below. $\pm$21,500 psi. This double plot is based on the microscopically observed mode of fracture. Whereas concurrent results were obtained at strain amplitudes above and below 0.005 in./in., there was a wide variation in the life of specimens tested at 0.005 in./in.

IV. Intermodal Transition Point

The different modes of fracture at high and low strain amplitudes cause a discontinuity in the ϵ-N curve at a strain level of 0.005 in./in. At this strain level, the short-lived specimen perhaps failed by the low-strain mode and the long-lived specimen by the high-strain mode as revealed by the figures 8 and 9. The crack in figure 9 appears to be more slip dependent and that in figure 8 is more intergranular. Hence both modes of fracture are possible at this strain level. A certain number of cracks in each specimen were chosen at random and the percentage of intergranular and transgranular cracks were roughly estimated in each case and plotted against strain amplitude (figure 12). The plot indicates an equal distribution of the two modes of fracture at the intermodal transition point.

B. FATIGUE OF BURNISHED COPPER

Surface cold working operations, such as surface rolling, burnishing and shot peening are well known for their pronounced enhancement of the fatigue life of metals and alloys.

I. The ϵ-N Curve

The effect of varying strain amplitude on the fatigue life of burnished copper specimens was studied and the ϵ-N curve for burnished copper is plotted in figure 13. Three distinct features are evident from the figure:

1. At low strain amplitudes, there is a vast improvement in the fatigue life of copper due to the burnishing operation. At a strain level of 0.004 in./in., the fatigue life of burnished copper was about 5 times that of annealed copper.
2. On the contrary, at higher strain levels, the burnished copper failed much earlier than annealed copper, the life being about half of that of annealed copper.
3. At the low strain level of 0.004 in./in., even though microcracks were seen at an early stage of 15,140,000 cycles (formation of microcracks coinciding with the automatic shut-off of the motor by the relay mechanism), the final fracture took place after enduring a large number of cycles, --- 20,163,000. This clearly suggests that there is a mechanism retarding the propagation of the microcracks in burnished copper.

II. The Fracture Mode

The failure mode in burnished copper was intergranular at all the strain levels investigated. This was only expected, because in the already cold worked layer of the burnished specimen, a complex dislocation network is formed, making further slip difficult. Consequently, a slip-dependent, transgranular fracture is not likely to take place in the surface worked metal at any stress level. The fracture could be due to stress concentrations at the dislocation pile-ups and intergranular, as in the case of high stress fatigue of annealed metal.

III. Effect of Burnishing

Surface cold working is a well known process for improving the fatigue life of machine parts such as axles and shafts (35). Roller burnishing is employed at the Ford Motor Company to improve the fatigue life of automobile front wheel spindles (40). For a long time, the beneficial effect of surface cold working was believed to be only due to the resulting residual compressive stresses. However, recently it has been shown by transmission electron microscopic technique, that very complex dislocation entanglements are produced by cold working a metal (41). Warke and Taylor (7) have shown that in shot peened copper subjected to fatigue loading at low strain, slip lines are entirely absent in the surface layers of the peened bars. In annealed copper, slip bands are seen on the surface and these bands are formed due to movement of dislocations over large distances near the surface. This is possible because there are no obstacles or entanglements to prevent the movement of dislocations. In shot peened bars, the absence of slip bands in the surface and sub-surface layers is due to the presence of complex dislocation forests which interfere with dislocation motion. Also in shot peened bars, at low strain level, even though a microcrack is formed at a very early stage, the complex dislocation forest retards the growth of the microcracks. A similar explanation holds good for the improved fatigue life of burnished copper. The burnishing operation makes the slip band formation at the surface difficult and makes the growth of microcracks very slow. At all strain amplitudes, the mode of fracture in burnished copper is intergranular. Figure 14 shows the fracture in one of the burnished specimens.

The shorter fatigue life of burnished copper at high strain levels than annealed copper is rather intriguing. A possible cause for this pronounced difference is perhaps due to the difference in the degree of strain hardening in annealed and burnished copper. The stress during fatigue loading of annealed copper goes partly to strain harden the grains and partly to cause fissures. Since the surface of burnished metal is already in a strain hardened condition, all the energy goes to the grains and this energy is released by the opening up of cracks at the grain boundaries. Further work is necessary to elucidate this mechanism.

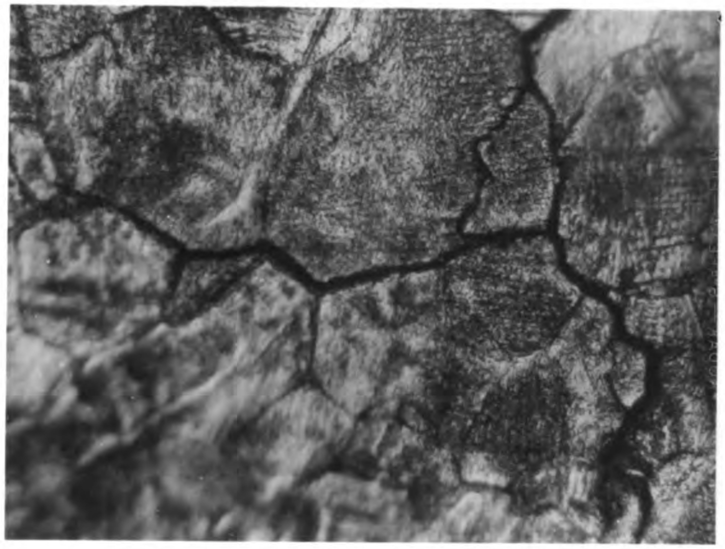

Figure 6. Fatigue crack in annealed copper at 0.007 in./in. shear strain. 570x.

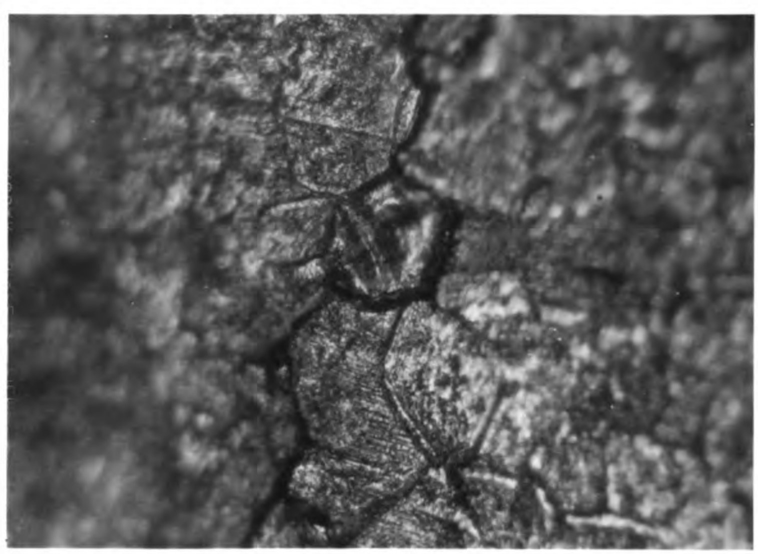

Figure 7. Fatigue crack in annealed copper at 0.006 in./in. strain. 570x.

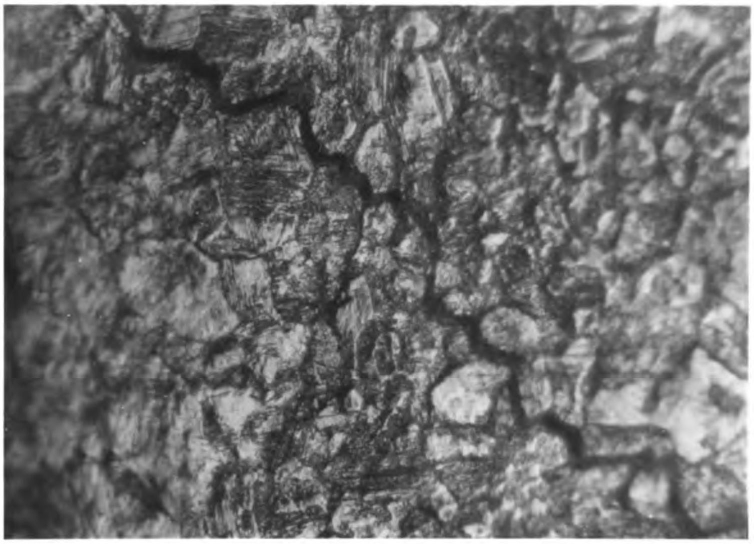

Figure 8. Fatigue crack in annealed copper at 0.005 in./in. strain, high strain mode. 570x.

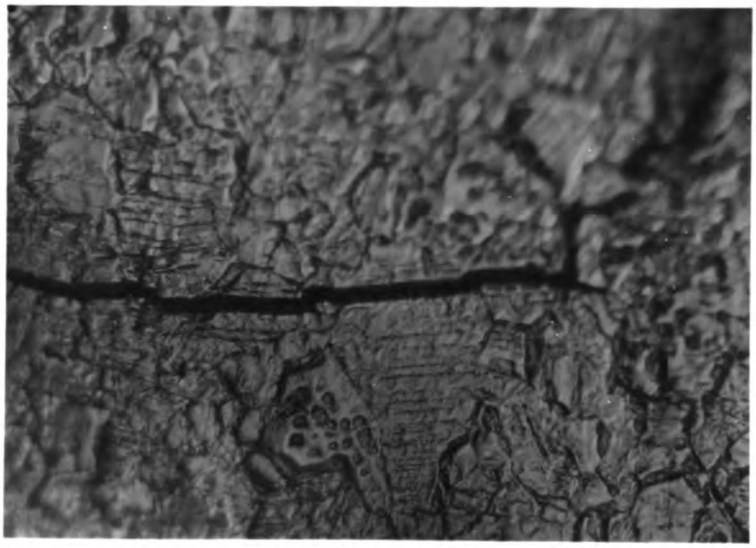

Figure 9. Fatigue crack in annealed copper at 0.005 in./in. strain, low strain mode. 570x.

Figure 10. Fatigue crack in annealed copper at 0.004 in./in. strain. 570x.

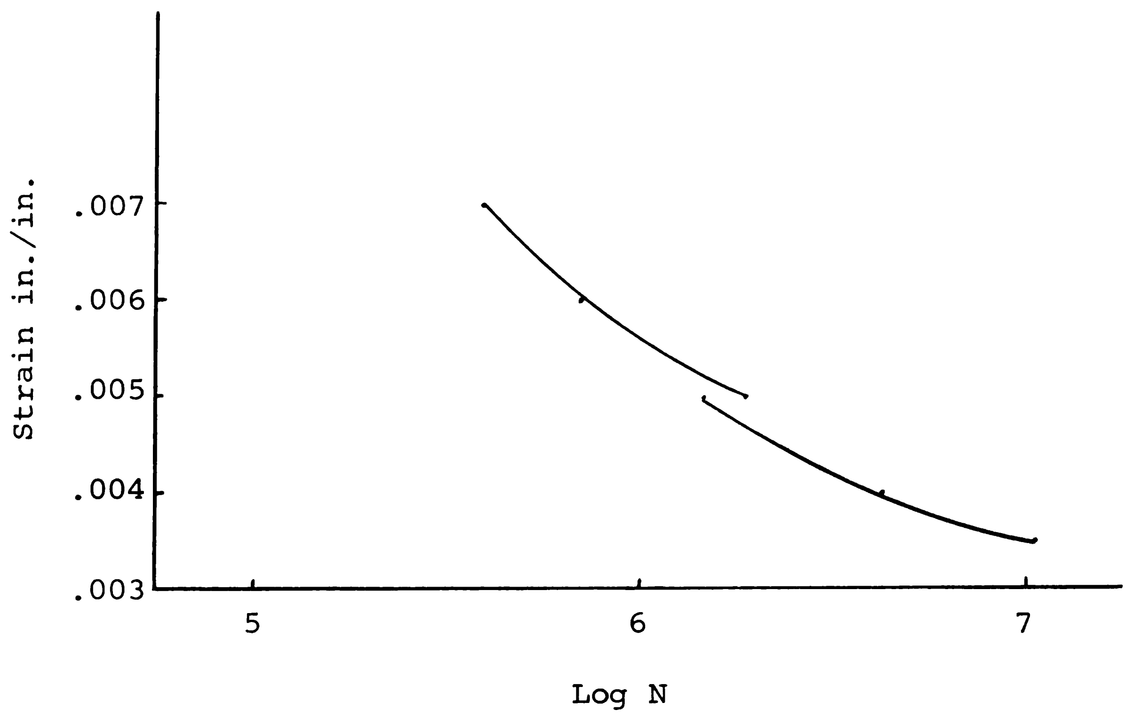

Figure 11. Strain amplitude vs. log of number of cycles for failure -- annealed copper.

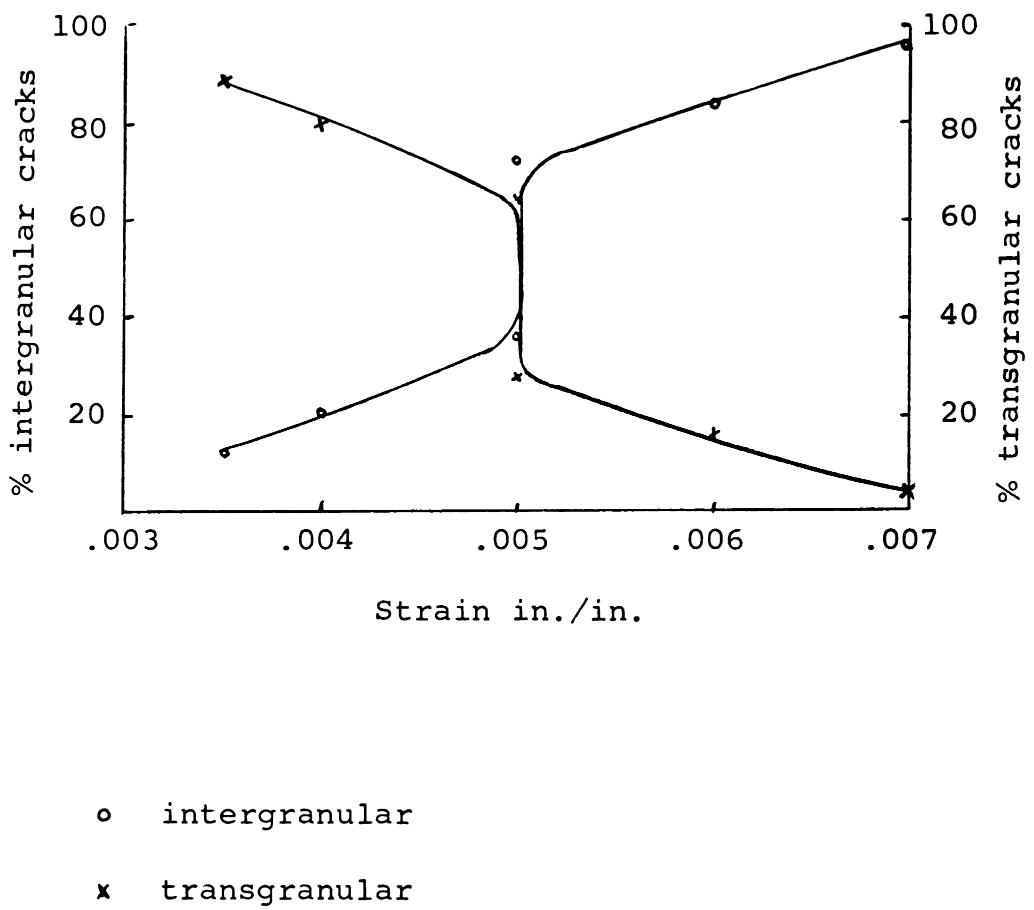

Figure 12. Mode of fracture vs. strain amplitude in annealed copper (schematic) (semi-quantitative)

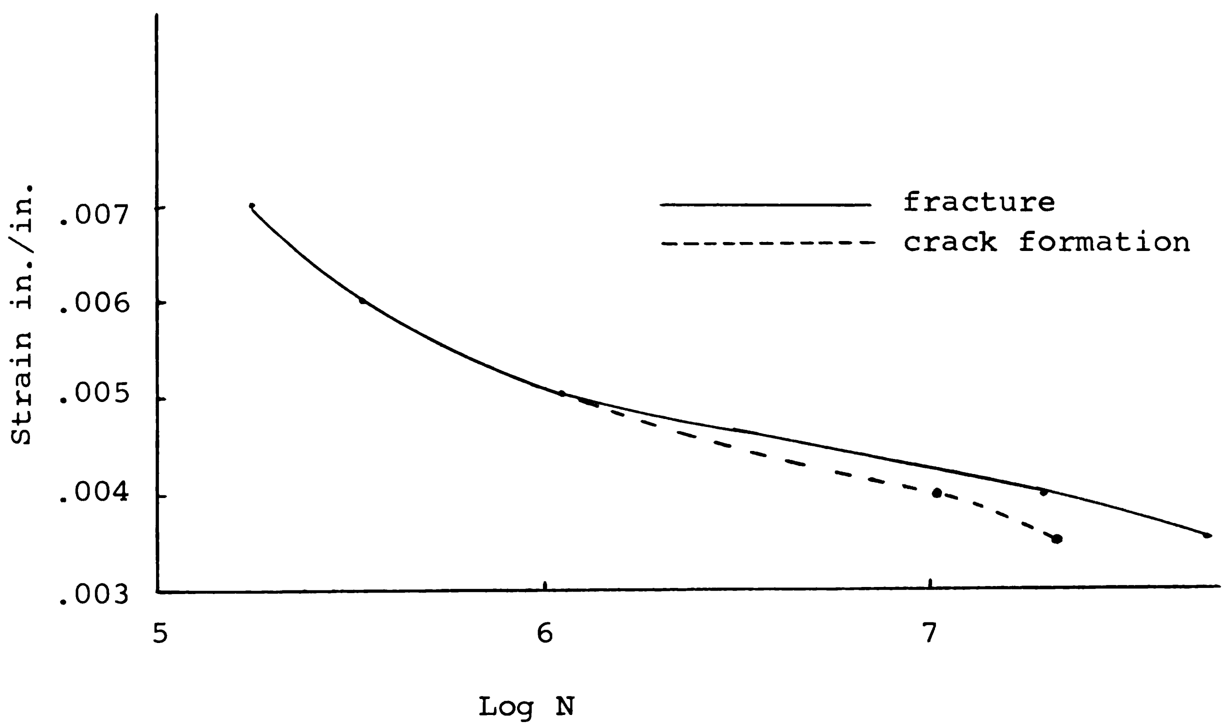

Figure 13. Strain amplitude vs. log of number of cycles for crack formation and fracture -- burnished copper.

Figure 14. Fatigue crack in burnished copper. 570x.

V. CONCLUSIONS

1. The mode of fatigue fracture in annealed copper is found to be definitely dependent on the strain amplitude.

2. Low strain amplitude leads to a transcrystalline fracture, generally following the slip line configuration.

3. At high strain levels, the fatigue fracture is intergranular.

4. There exists an intermediate strain level which may be termed "Intermodal Transition Strain" at which both types of fracture are possible.

5. The constant strain amplitude fatigue fracture of burnished copper is intergranular at all strain levels.

6. The number of cycles between the appearance of microcracks and failure in burnished copper, at low strains, was very large and the corresponding number for annealed copper was negligible.

7. At high strain levels, the fatigue life of burnished copper was shorter than that of annealed copper. Further work needs to be done to elucidate a mechanism for this.

VI. RECOMMENDATIONS

It may be worthwhile investigating the following aspects:

1. At low strains, the fracture is transgranular and at high strains, it is intergranular. Specimens of large grain size and hence small grain boundary areas may be tested in fatigue at high strains and specimens of fine grain size tested at low strains, to find the effect of grain size on fatigue life for various strain amplitudes.

2. At low strains, introduction of a complicated dislocation network in the surface layers by cold working, leads to an enhancement of the fatigue life. Other obstacles, such as dispersed oxides and foreign atoms may be introduced by internal oxidation, diffusion alloying etc. and their effect on fatigue life may be studied.

3. Study of the dislocation movements during fatigue loading at various strain levels may be made by the transmission electron microscopic technique.

VII. REFERENCES

1. Sines, G. and Waisman, J. L., Metal Fatigue, McGraw-Hill Book Co., Inc., New York, 1959, 3.

2. Gensamer, M., Fatigue and Fracture of Metals, The Technology Press of M.I.T. and John Wiley & Sons, Inc., New York, 1952, 1.

3. Waisman, J. L., Metal Fatigue, McGraw-Hill Book Co., Inc., New York, 1959, 7.

4. Prevention of the Failure of Metals under Repeated Stress, Battelle Memorial Institute Staff, John Wiley & Sons, Inc., New York, 1941.

5. Peterson, R. E., ASTM Bull., <u>205</u>, 50 (1950).

6. Porter, J. and Levy, J. C., J. Inst. Met., <u>89</u>, 86(1960-61).

7. Warke, W. R. and Taylor, W. E. (in press).

8. Fuchs, H. O., Metal Fatigue, McGraw-Hill Book Co., Inc., New York, 1959, 197.

9. Thompson, N., Fatigue in Aircraft Structures, Freudenthal, A. M., Academic Press Inc., New York, 1956, 43.

10. Ewing, J. A. and Humfrey, J. W. C., Phil. Trans. Royal Soc., <u>200</u> A, 241 (1903).

11. Gough, H. J., Proc. ASTM, <u>33</u>, II, 3 (1933).

12. Orowan, E., Proc. Royal Soc., <u>171</u> A, 79 (1939).

13. Polakowski, N. H. and Palchoudhuri, A., Proc. ASTM, <u>54</u>, 701 (1954).

14. Wood, W. A. and Segall, R. L., J. Inst. Met., <u>86</u>, 225 (1957-58).

15. Kemsley, D. S., ibid., <u>87</u>, 10 (1958-59).

16. Alden, T. H. and Backofen, W. A., Acta Met., <u>9</u>, 352 (1961).

17. Bullen, F. P., Head, A. K. and Wood, W. A., Proc. Roy. Soc., 216 A, 332 (1953).

18. Wood, W. A. and Davies, R. B., ibid., 220 A, 255 (1953).

19. Forsyth, P. J. E., Fatigue in Aircraft Structures, Freudenthal, A. M., Academic Press Inc., New York, 1956, 20.

20. Hempel, M. R., Fracture, The Technology Press of M.I.T. and John Wiley & Sons, Inc., New York, 1959, 376.

21. Bendler, H. M. and Wood, W. A., Trans. AIME, 224, 180 (1962).

22. Thompson, N., Fracture, The Technology Press of M.I.T. and John Wiley & Sons, Inc., New York, 1959, 354.

23. Wood, W. A. Phil. Mag., 3, Ser. 8, 692 (1958).

24. Wood, W. A., Fracture, The Technology Press of M.I.T. and John Wiley & Sons, Inc., New York, 1959, 412.

25. Bendler, H. A. and Wood, W. A., Acta Met., 8, 402 (1960).

26. Wood, W. A., Fatigue in Aircraft Structures, Freudenthal, A. M., Academic Press Inc., New York, 1956, 1.

27. Cottrell, A. H., Dislocations and Plastic Flow in Crystals, Oxford University Press, 1953.

28. Kemsley, D. S., J. Inst. Metals, 85, 153 (1956-57).

29. Strom, B., ibid., 83, 530 (1954-55).

30. Kemsley, D. S., ibid., 85, 420 (1956-57).

31. Forsyth, P. J. E., ibid., 80, 181 (1951-52).

32. Valluri, S. R., Aerospace Engg., 20 (10), 18 (1961).

33. Rosenthal, D., Metal Fatigue, McGraw-Hill Book Co., Inc., New York, 1959, 170.

34. Brookman, J. G. and Kiddle, L., Symposium on the Failure of Metals by Fatigue, Melbourne University Press, 1946, 395.

35. Horger, O. J., Surface Stressing of Metals, American Society for Metals, Cleveland, 1947, 85.

36. Thompson, N., Wadsworth, N. J. and Louat, N., Phil. Mag., 1 (viii), 113 (1956).

37. Stroh, A. N., Proc. Roy. Soc., 232A, 548 (1955).

38. Kemsley, D. S., Nature, 178, 653 (1956).

39. Louat, N., Ph.D. Thesis, University of Bristol, cited by Kemsley, D. S., J. Inst. Metals, 85, 153 (1956-57).

40. Metlfax, 5 (11), 45 (1961).

41. Berghezan, A. and Fourdeux, A., J. Appl. Phys., 30, 1913 (1959).