# A STUDY OF THE EFFECT OF SURFACE DEFORMATION ON THE FATIGUE PROCESS IN COPPER

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# ABSTRACT

## A STUDY OF THE EFFECT OF SURFACE DEFORMATION ON THE FATIGUE PROCESS IN COPPER

By William R. Warke

The effect of surface working as accomplished by shot peening on the fatigue process in copper has been studied. It is found that surface working has a profound effect on the slip band distribution, inhibiting slip in the deformed layer. Since slip band formation at free surfaces leads directly to fatigue failure, inhibiting such formation causes increased life. Shot peening produces two effects in the surface layer: it is left with residual compressive stresses and it is cold worked. It is not clear which of these two produces the above result, further work being needed to determine this. There are also indications that crack propagation is slowed in the surface worked bars.

# A STUDY OF THE EFFECT OF SURFACE DEFORMATION ON THE FATIGUE PROCESS IN COPPER

By

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# I. INTRODUCTION

For over a century, the problem of metal fatigue has been a source of consternation to engineers and designers as well as scientists (1). As long ago as 1849, discussions were held by mechanical engineers in England with reference to the premature failure of axles in railroad rolling stock and even today, the vast majority of service failures are attributed to fatigue. In the period around 1860, Wöhler ran experiments on this phenomenon and is credited as being the first to consider the problem from an organized, experimental point of view. It was at about this time that the term fatigue was first applied to the apparently brittle premature failure of machine parts under repetitive loads.

Investigations into the problem of metal fatigue are carried on at three levels. The first is the macroscopic level dealt with by Wöhler and countless others since his day. These investigations are typified by the well known S-N curves which give the number of applications of any given stress required to produce fracture. At this level, almost every conceivable variable has been studied. These include metallurgical parameters such as microstructure and mechanical factors such as notch effects. In these studies, the interest is in complete failure and the number of cycles to produce it, there is no concern with the mechanism responsible.

Near the end of the last century, the metallurgical microscope was developed and soon afterwards was applied to the fatigue problem. (1) Thus, work by Ewing, Rosenhain, and others around the turn of the century was the start of investigations on the second level. Work on this level is typified by observation through the microscope of the progress of fatigue damage as a function of the number of cycles of application of a stress or strain. From the very first investigation it was noted that slip concentrated into bands under cyclic loading and that the final fracture started in these bands. As a result of this type of study, much has been learned about the fatigue process with exceptional advances being made in the past decade.

The third level of study had its beginning only within the last few years. It consists of devising a mechanism for the processes occurring on an atomistic level in the presence of alternating stresses or strains. It deals with the motion of dislocations, their generation and interactions with each other and with grain boundaries, free surfaces and other obstacles which may be present. A valid theory must be consistent with observations and existing knowledge on all three levels. In the next section, various recent theories of fatigue will be discussed in this light.

The many years of investigations on the macroscopic level, coupled with studies of service failures, has led to a body of empirical knowledge regarding practices and treatments which are either harmful or helpful with regards to fatigue life (2). For instance, some harmful factors are designing with sharp reentrant angles or fillets, a corrosive environment and poor surface finish. Some beneficial factors are a polished surface, compressive residual stresses in the surface, designing with smooth transitions in section and local working of the surface layers.

In view of the recent advances in the microscopic aspects of fatigue, it is of interest to begin to correlate the known factors at the macroscopic level with their corresponding effects on the fatigue process on the microscopic level. In this study, an attempt has been made to initiate such an investigation for the case of surface cold working. Surface cold working has a dual effect. First of all, it produces work hardening of the affected layer. Secondly, it produces residual compressive stresses in the surface. There are several commercial processes for surface cold working. Among these are shot peening, surface rolling, and burnishing. In this investigation, shot peening was chosen as the method to be used.

## II. BACKGROUND AND THEORY

In the following discussion, the fatigue process in "simple" fatigue will be considered. That is we will be dealing with fatigue of pure metals, for the most part, as polished bars subjected to completely reversed stress or strain. External stress raisers or any other external variable will not be considered. Thus, the fatigue under discussion will be a basic property of the material itself. The first part of the discussion will center on the experimental evidence at the microscopic level regarding the fatigue process. Then some of the theories used to explain the observations will be covered.

### A. Experimental

The first investigators to study the surface appearance of specimens under fatigue loading were probably Ewing and Humphrey in 1903 (3). Slip lines in Swedish iron were observed to form bands of closely spaced lines. As cycling continued, these grew wider and more intense. Fatigue cracks were formed in these slip bands and these then propagated to produce fracture. These same observations have been made using a wide variety of metals, since then and more modern investigations have been for the most part using more refined techniques to study the process.

The fatigue process can be divided into three stages. These stages are most evident in tests which involve a constant strain amplitude rather than a constant stress. Some workers prefer to maintain a constant plastic component of strain while others maintain the total strain at a constant level. The first observable stage consists of work hardening of annealed material or softening of cold worked material to a stress level and hardness characteristic of the strain amplitude employed. For instance, Coffin and Tavernelli show that for a three percent plastic strain applied to annealed OFHC copper, the initial stress range will be 35,000 psi, while material having been cold worked thirty-three percent showed an initial stress range of 100,000 psi (4). After approximately one hundred cycles of strain, both showed a stress range of about 85,000 psi. Kemsley shows that two sets of copper specimens having respective hardnesses of 38 DPH in the annealed condition and 95 DPH in the cold worked condition, both had hardnesses of 75 DPH after cyclic loading (5). Further work on strain hardening and softening of metals produced by cyclic strain has been done by Wood and Segall (6) and by Polakowski and Polchoudhuri (7). These workers found changes in proof stress on the next cycle and in compressive yield strength and Vickers hardness, respectively. In any case, the rate of work hardening for cyclic deformation was less than that for unidirectional strain in both single- and poly-crystalline specimens.

During this stage, slip is disturbed uniformly throughout the grains, there is no visible concentration of slip. Also, X-ray back reflection photographs of annealed metals show that fragmentation of the grains is taking place (8, 9). The sharp spots characteristic of annealed metal tend to broaden along Debye rings. At high deflections continuous rings will be formed. Also, the broad, fuzzy rings characteristic of cold worked metal will sharpen under the influence of a cyclic deformation. However, for any given strain amplitude, the rings for annealed and cold worked material will not become identical although they tend in that direction (5). Re-annealing at this stage of originally annealed material causes it to revert to its original condition, thereby extending the life (10).

After a number of cycles, which is a function of the strain amplitude employed, work hardening ceases and the second stage, termed saturation by some, begins (10). The number of cycles required to reach saturation increases with increasing amplitude (11). At high amplitudes, fracture may occur before saturation is attained. The second stage is typified by non-hardening plastic deformation. No further changes occur in the X-ray diffraction patterns from this point on (12). It is during this stage that slip clusters in closely spaced lines causing the appearance of bands. Once this occurs, annealing will have no statistically significant effect on the remaining life (13).

Much effort has been expended in the study of the formation and nature of these slip bands, since the fracture was known to originate in them. They have been studied in a wide variety of materials including copper (11, 12, 14, 15, 16, 18), aluminum (10, 12, 16), brass (11, 12), nickel (12), gold (17), and various steels (18, 19, 20, 21). It is found that they develop certain characteristics almost immediately which eventually result in their being the source of fracture. The material in the slip band becomes either raised or lowered with respect to the undisturbed bulk of the grains. These surface distortions have become known as extrusions and intrusions.

Various techniques have been used to study the extrusions and intrusions. Several investigations have used electron microscopy (5, 20, 21). By using a positive replica and by appropriate shadowing, they have studied the topography of the bands and find microcracks at a very early stage. Thompson, Wadsworth and Louat removed a few microns of metal from the surface by electrolytic polishing (18, 22). When this is done at only 5% of the total life, it is found that most of the slip bands are removed. However, some remain in evidence and are termed persistent slip bands. Continued electro-polishing showed that these can be as much as thirty microns deep after 25% of the total life. If a tensile force is applied, those persistent slip bands involving more than one grain open up, while those within one grain only broaden slightly. These are differentiated by terming short ones fissures while the ones in two or more grains are called cracks.

Kemsley found that by sectioning and proper etching of fatigued copper bars, he could reveal slip bands in the interior of the bar as well as at the surface (15). High stress produced bands which were shorter and less numerous. When persistent slip bands were present, they appeared to be associated with the etched up markings. Wood modified this technique somewhat (11, 23, 24). He realized that if the polish plane intersected the surface at a low angle, the topography of the surface would be magnified somewhat. This coupled with the optical magnification would allow close inspection of the nature of the slip bands at the surface. By polishing a narrow longitudinal flat on the cylindrical specimen, he obtained mechanical magnifications of about twenty times giving a total magnification of twenty to thirty thousand times. Electron microscopy has also been used to give even higher magnifications (25). The results obtained show that after as little as 10% of the total life, the slip bands develop distinctive profiles, being either sharp notches, peaks or intermediate combinations of the two. As cycling continues, the fissures can be seen to form in those etched up slip bands which have a notch type cross section. These than become cracks and propagate by various means across the grains. Higher amplitudes produce shorter bands and at sufficiently high amplitude, none appear indicating a change of mechanism. Annealing prevents the bands from etching up but does not heal the fissures or change the topography. Experiments on gold show that annealing for 96 hours at 950$^{\circ}$ C is required to heal the fatigue damage any significant amount (17).

Ebner and Backofen revealed the slip band topography by placing a Knoop Hardness Test indentation across the slip bands (14). Since the angle between the long faces of the indenter is 170$^{\circ}$ 30', a low angle of intersection between the indentation and the surface is obtained. They found the same effects at the surface as Wood. That is the slip bands were either notched, peaked or a combination of the two, at the surface. However, this technique gives no information about the underlying material.

Alden and Backofen measured the rate of growth of the extrusions and found that they develop at the rate of one to two microns per thousand cycles which corresponds to a net movement of three to six dislocations per cycle (10). These results were for a total strain amplitude of two-tenths of one percent.

The final stage of fatigue failure is crack propagation. At low amplitudes, the cracks are observed to follow the slip bands. This results in the usual transcrystalline fracture observed for fatigue at relatively low stresses and strains. At high amplitudes of stress and strain, the fracture is both intercrystalline and transcrystalline. When crossing a grain, at high amplitudes, there is no apparent dependence of the path on slip bands. During this portion of the test, the stress required to produce a constant strain had been observed to both increase and decrease depending on the material and conditions of the test.

To summarize, then, the fatigue process consists of three stages; work hardening or softening as the case may be, crack formation and crack propagation leading to failure. Plastic flow occurs at all stages if failure is to occur but need be accompanied by hardening only in the first stage. X-ray and metallographic studies show slip to be uniformly distributed in the first stage and concentrated in the later stages.

Second stage slip is clustered in bands and produces extrusion and intrusion of material in the band leading to notch-peak cross sections. Annealing will repair damage during the first stage but not during the later stages. There is evidence that at high amplitudes, the second stage as has been defined here does not occur and the crack forms by another mechanism. A different mechanism of crack propagation at high amplitudes is also indicated.

## B. Theory

One widely held theory of fatigue fracture was originally proposed by Gough but is generally attributed to Orowan since he refined and expanded Gough's ideas to a more precise theory (26, 27). This theory is sometimes called the strain hardening theory of fatigue. In a ductile material, yielding will occur locally when the stress at some point in the material reaches a critical value. This occurs because of inhomogeneities such as grain boundaries, inclusions or other defects in the material. Fracture will occur at some critical value of stress and strain which is peculiar to the particular metal. As the material is cyclicly strained, yielding occurs at the local stress peaks and work hardening occurs. As the strain is reversed, the material continues to work harden and the strain to be considered is the algebraic sum of the individual strains. As the material hardens, the local strains decrease as the terms of a geometric series, so that the sum is always finite. If the sum is less than the critical value, the range of stress is safe, if larger, it is unsafe and results in fracture at that point. There are several weaknesses in this theory. First of all, fracture occurs at much lower stresses than the fracture stress of the material. It is hard to conceive that local fluctuations could make up the difference. Secondly, the algebraic sum of the individual strains is many times more than the fracture strain in a unidirectional test. Also, this theory is not in keeping with the observed non-hardening second stage of the fatigue process. Finally, it does not account for the fact that the cracks originate at the free surface in all cases of simple fatigue.

Another theory, due to Machlin, assumes that there are pre-existing cracks in the material (28). Due to stress concentrations at the ends of the crack, dislocations will be generated there. As a result, the size of the crack will grow until fracture occurs. This theory fails to account for the first and second stages observed in fatigue testing. Also, it does not account for the fracture starting from the surface.

There are several theories which attempt to account for the three stage process for fatigue which has been described. First, an explanation must be made of the formation of fissures in the slip bands and second, an explanation is needed of why slip is concentrated in bands. Several of these theories center around assuming and explaining the loss of cohesion across the slip band. Fruedenthal proposes that a certain proportion of the bonds broken at the surface on a slip plane by strain in one direction will not reheal upon reverse straining (29). This occurs over many cycles and thus eventually leads to loss of cohesion across the slip plane. Others arrive at a similar conclusion by the fact that when a moving edge dislocation passes through a forest of screw dislocations, jogs are formed. As the dislocation continues to move, the jogs result in the generation of vacant lattice sites. Since the slip is concentrated in bands, these will be regions of high vacancy concentration. Under the influence of the cyclic strain, and by diffusion, these then coalesce to form voids and fissures (7) which grow and connect so that a crack forms in the slip band. Another variation on this idea is that gases such as oxygen are adsorbed on the surfaces of the slip band preventing cohesion and producing an interface (18). These theories explain some experimental observations but cannot be the dominant mechanism. Vacancy production explains the etching up of slip bands both near the surface and in the interior of the bars and also the disappearance of etched up bands upon annealing. This would occur by diffusion to nearby dislocations resulting in climb.

Also, it has been shown that protecting the fatigue test bar from oxygen or corrosive media by inert gas or other means, extends the life (18). However, removal of the etched up bands by an anneal does not change the total life nor does an inert atmosphere prevent failure. Furthermore, McCammon and Rosenberg and others have obtained fatigue failures at temperatures as low as 4.2°K (30). Diffusion would be non-existent at this temperature for all practical purposes. It was found that the ratio of fatigue limit to fracture stress remained more or less constant over the range from 4.2 to 298°K for the materials studied. This indicates that the variation with temperature of the fatigue limit is probably related to the variation in the flow stress.

Mott has proposed another theory in which the slip band is considered to become extremely disordered due to increased vacancy concentration and other results of slip being concentrated (31). This results in a volume expansion. Eventually the disorder results in recrystallization accompanied by contraction in a thin region. This opens up the crack. Fatigue fractures at low temperatures also invalidate this theory.

Recently, Wood has proposed a theory which seems to fit the experimental evidence to a greater degree than the aforementioned theories (11, 23, 24). In this theory, fracture at low amplitudes is considered to be a direct consequence of the geometry of the slip bands at the surface. At higher amplitudes, a theory such as Orowan's is acknowledged to be probably applicable. That is, there are two mechanisms active, the one applicable depending on the amplitude of stress or strain. The fact that during the second stage of fatigue, intrusion and extrusion occurred had long been recognized as mentioned in the last section. However, this had never been considered the cause of failure. Wood showed by his taper section technique that sharp notches developed at as little as ten percent of the total life. These result from the fact that slip may occur on one plane in one direction but may not occur on the same plane in the opposite direction but rather on a nearby plane. As a result, the material between is left either raised or lowered with respect to the surface. When this is repeated over a thousand or so cycles of even a small strain, a deep, sharp notch is obtained. Once this occurs, it is impossible to heal the damage since by then the notch or peak, as the case may be, is a self perpetuating stress raiser. Thus, although annealing may cause diffusion of the vacancies and dislocation climb to remove some of the slip band damage, it can not in a reasonable length of time fill in the notches or level the peaks by bulk diffusion or sintering. As soon as cycling is begun again, the notch or peak causes slip to occur in its immediate vicinity and it continues to grow as a result. However, the vacancy concentration may make it easier for the crack to propagate along slip bands and thus contribute somewhat to the process. This picture adequately fulfills the requirement of explaining the formation or nucleation of cracks or fissures in slip bands. Persistent slip bands would be simply deep notches which form as a direct result of the intersection of slip with the free surface. The crack growing from the free surface and the temperature dependence are also explained easily. Slip bands also form in the interior but have no effect due to the lack of a free surface.

The problem of explaining why slip concentrates into bands during the second stage of fatigue still remains to be solved. Furthermore, the notch-peak cross sections must be explained as well as the absence of work hardening. To do this, dislocation theory must be applied. At this level, it must be remembered that we are dealing with pure theory, with very little basis on observation. All of the hypotheses have points of weakness, and there is little to cause one to be more favorably viewed than another.

One attempt along these lines was made by Cottrell and Hull (32). They envisioned two sources on intersecting slip planes operating alternately on each half cycle. As a result, each source is displaced slightly from its position with respect to some point on the surface on the previous half cycle. This causes an extrusion to form at the surface due to one source and an intrusion due to the other in one full cycle. Although such intersecting slip has been observed by Wood on taper sections, it did not appear to be the rule. Also, it fails to explain the non-hardening aspect of the second stage and the formation of etched up bands in the interior of the material.

Mott has proposed a theory which is based on the concept of cross slip of screw dislocations to explain slip band formation (33). He claims that the first stage is due to work hardening and operation of dislocation sources. The second stage begins when the material work hardens to the point where the applied stress can no longer cause sources to operate but can still move existing dislocations. If there are dislocations in the band which are free to move, extrusions and intrusions can form in the following manner. The type of dislocation responsible is a screw dislocation with one end at the free surface with a component of the Burgers vector perpendicular to the surface. If this dislocation, then, follows a closed path, resulting from cross slip at either end of the cycle, extrusion or intrusion will occur to the extent of the number of circuits times the normal component of the Burgers vector. This process would then produce a cavity at the base of the extrusion. However, such cavities have not been observed. Also, this too has difficulty explaining the slip bands occurring in grains not at the free surface.

Two similar theories are proposed by Thompson and Nagai (22, 34). These require the operation of dislocation sources to explain the formations. As the loops produced by the source expand toward the surface, they cut through a "forest" of screw dislocations on intersecting planes. This produces jogs in the loop, whose motion then generates vacancies. This accounts for the etching up of slip bands. Thompson then proposes that the dislocations which make up the forest will bow due to the applied stress. The direction of bowing will reverse when the stress is reversed. Thus the jogs in the dislocations arriving at the surface during the negative half cycle will be displaced from those of the positive half cycle. This will result in imperfect cancelling out of the prior damage at the surface, leaving small intrusions and extrusions.

Nagai, on the other hand, makes assumptions regarding the form and action of the source. He envisions a length of edge dislocation pinned at each end by screw dislocations. As the source operates, the slip plane is moved up or down an amount equal to the Burgers vector of the screws. On the reverse half cycle, the slip plane moves in the opposite direction. This results also in imperfect cancellation of previous damage when jogs reach the surface. However, it is doubtful that a source with this configuration could exist since the sum of the Burgers vectors at a node must be zero. This theory does have the advantage though of predicting a band of some width which is a fault of the previous theory. Furthermore, neither of these theories explains the non-hardening aspects of the problem. It is easy to see that this phase of the problem is far from solved and a reasonable theory of initial intrusion-extrusion formation is still to be found.

# III. EXTENSION OF FATIGUE LIFE

As stated in the introduction, methods of extending fatigue life have long been known. This is both from the standpoint of things to avoid as well as positive measures to take. These discoveries were made on the macroscopic level and empirical rules of this nature are well established. However, now that the process by which fatigue damage occurs is better understood, it is of interest to see how these methods of extension of life affect the microscopic processes. Very little work has been undertaken along these lines.

One such investigation was included in the work of Thompson, Wadsworth and Louat (18). They studied the effect of exclusion of oxygen from the surface of copper fatigue bars. This had been found to increase the endurance limit of copper from 1.02 to 1.13 times depending on the type of copper when tested in vacuum. It was found that exclusion of oxygen had no effect on the rate of formation of slip bands but that the rate of spread of persistent slip bands was lower or cracks took longer to spread. That is, oxygen caused the cracks to propagate more quickly, but the cracks formed at the surface in any case. On the other hand, Hempel found no difference in either total life or slip band formation when a plain low carbon steel was tested in oxygen, nitrogen, hydrogen, argon or air (21).

Alden and Backofen studied the effect of anodic films on the fatigue process in single crystals of aluminum (10). They found that a thick, unbroken film prevents the formation of slip bands and thereby the formation of cracks. Any crack in the film becomes the source of failure while material under the intact film remains uncracked. The film serves to block the formation of slip steps and notch-peak topography. This is further evidence that slip band cracking is the direct result of active slip in the presence of a free surface.

Several studies have been made concerning the effect of removal of the surface layer after various lengths of time (13, 17, 18). It is found that this effectively extends fatigue life and if periodically done may extend it indefinitely. Removal may be done either by machining or by electropolishing. In either case, the life is extended if enough material is removed to get past all surface cracks and persistent slip bands. However, if insufficient material is removed by electropolishing, the effect may be to shorten life. This offers quite conclusive proof that fatigue damage is limited to the surface layers apart from the annealing experiments already described. This is the extent to which work of this type has been done.

Surface working has been recognized for some time as an effective and practical means of increasing fatigue life of machine parts, especially for applications where the load is either in bending or torsion (35, 36, 37). Surface rolling is used for cylindrical parts such as shafts and locomotive crank pins. This is usually done by a three roll device with a means for controlling the load on the rollers, mounted on the carriage of a lathe and passed along the part by the lathe feed. Surface rolling is also used for fillets. A narrow roller is forced into the fillet and passed back and forth across it. Burnishing is also used for surface working (38). An additional gain is found in the high surface finish obtainable. These three methods, although very useful, are limited in application to simple shapes. Another method of surface working which is applicable to complex shapes as well is shot peening. A stream of small steel or cast iron shot is directed at high velocity onto the surface of the parts. Shot sizes range from .007" to .175" and velocities of 100 to 200 ft/sec are used. The intensity of peening is measured by the deflection of a standard strip called an Almen strip which is peened on one side. This process is used widely in the spring industry for both coil and leaf springs and is very effective in improving fatigue properties. Considerable data are available in the literature regarding the life of particular machine parts as well as the shift in the S-N curve of various materials resulting from shot peening.

When the shot strikes the surface upon which it is directed, the material is compressed in the normal direction and extended radially in a layer about ten thousandths of an inch thick. This results in plastic deformation of this layer with its attendant work hardening and permanent change of shape. When the shot removes, the layer is left in a compressed state in the radial directions. This is due to the fact that the surface layer which is stretched by the shot remains continuous with the underlying material which has not changed its shape. In order to retain this continuity, the stretched portion must be compressed to approximately its original dimensions. Residual compressive stresses can be obtained in this manner which approach the yield strength of the material. The extent of work hardening of the surface is appreciable as well. For example, a steel having an initial hardness of 350 DPN may have the hardness of the surface increased to 400 DPN by shot peening.

When considering the effect of surface working on fatigue, the reverse effects cannot be neglected. That is, consideration must be given to the effect of fatigue loading on the residual stress distribution and the intensity of cold work in the surface layer. Some discussion has already been devoted to the effect of cyclic loading on uniformly cold worked material and the same results can be expected in the worked surface of a peened bar. It will be remembered though, that cyclic loading failed to soften cold worked metal to the same state as annealed metal is hardened as measured by X-ray methods. So, cyclic loading can be expected to effect some recovery and softening of the surface layer.

Also, cyclic loading acts to reduce the intensity of the residual stresses and level out the stress distributions. However, in this case too, the effect is incomplete and some stresses remain after millions of cycles. This is true regardless of whether the source of the residual stress is thermal or mechanical in nature. Much work along this line has been done by Bühler and Buchholtz (36).

Even very recently, the beneficial effect of surface working was attributed solely to the resulting residual stress with the accompanying cold work taking a minor place. This may not be the case. Some have gone so far as to say that a tensile stress is necessary for fatigue failure and if the compressive residual stress is large enough, failure is prevented. From the previous discussion it would seem that a more likely statement would be that the presence of a fluctuating shear stress is all that is needed for initiation of a fatigue failure.

# IV. EXPERIMENTAL

The material used in this investigation was oxygen free high conductivity copper. The nominal composition of this material is 99.98% copper. It was obtained from S. Mostovoy of the Illinois Institute of Technology as 5/8" hexagonal bars. These were cold drawn to 1/2" round giving a final hardness of 92 Rockwell F. Standard R.R. Moore fatigue specimens were then machined from the bar. A gauge length with a nominal diameter of .300 inches and length of .750 inches was provided. The bars were mounted in a lathe and the gauge length was polished with successively finer grades of emery paper down to 3/0 grade. This was followed by an annealing operation. Annealing was done in a resistance heated tube furnace with an atmosphere of argon flowing through the tube at all times. A thermocouple was kept in the immediate vicinity of the specimen during the annealing cycle. A specimen was placed in a boat which was then placed just inside of the tube which was then covered. After five minutes which were allowed to purge oxygen from the system by the flowing argon, the boat was slid to the hot zone of the furnace by a rod extending through the tube covering. The furnace, having been preheated to the annealing temperature of 1250$^{o}$F, was cooled considerably by the introduction of the specimen and required an additional five minutes to return to temperature. After holding at 1250°F for ten minutes, the specimen was quenched in water. This produced a hardness of 53 Rockwell F.

After annealing, the bars were electropolished in a two to one solution of methyl alcohol and nitric acid. Approximately five minutes with a current of four amperes were required to remove the fine scratches remaining from the final emery paper. This left a scratch free polished surface with only a little grain contrast and some surface oxidation visible under the microscope. Several samples were retained in this condition and were placed in a desiccator to prevent excessive oxidation.

The remaining samples were shot peened. This was done at the General Motors Research Laboratory through the courtesy of Dr. Douglas Harvey. The peening was done with #54 cut wire shot to an intensity of .0098 Almen A strip. That is, a strip 3" by 3/4" by .050", having a hardness of 47 Rockwell C, when peened on one side, only gave a deflection of .0098" when measured over a 1.25" span by a dial gauge.

Testing was originally intended to be carried out on R. R. Moore fatigue machines but due to severe mechanical difficulties their continued use was prohibited. It seemed that due to the softness of the material, and the high rotational speed of the machine, any small inaccuracies in machining or alignment gave a disproportionately large force on the specimen and machine. As a result, several specimens were ruined and there was some minor damage to the machine.

It was then decided to construct a controlled deformation machine. The machine, shown in Figures 1 and 2, is a constant strain amplitude torsional fatigue machine.

Since the specimens were already made with the appropriate taper to fit into an R. R. Moore machine, it was necessary to design around the specimen holder shafts from such a machine. Two such shafts were mounted in an arbor, the rear one being held by a clamp and simply slid into the arbor. The front shaft was mounted on preloaded taper bearings and driven by the lever arm. The specimen was mounted by removing the rear shaft, mounting in the front shaft, replacing the rear shaft and mounting therein and finally clamping the rear shaft. Care in this operation had to be exercised to minimize stressing the specimen during the mounting and clamping operations. The machine was driven by a General Electric, 1/4 horsepower, 3450 rpm motor.

A $\pm 1^{\circ}$ motion, imparted to the lever arm by an eccentric cam at the motor is transmitted to the specimen producing a calculated shear strain of .0035 in/in. The torque on the specimen is measured by strain gauges mounted on the constriction on the lever arm and can be converted directly into shear stress. The output from the strain gauges was amplified by an Ellis Bridge Amplifier and then displayed on a Techtronik Model 532 Oscilloscope. This system was calibrated periodically between the tests by removing the motor and hanging weights from the lever arm at a point corresponding to the centerline of the motor shaft, 5 inches from the specimen center line.

The slip process during fatigue was studied by microscopic examination. In the case of polished bars, observations could be made directly on the surface. However, due to surface roughness from the peening operation, this could not be done on peened bars. Also, it is desirable to see the events taking place below the surface of the bars. For these reasons a taper sectioning technique similar to that developed by Wood was used. In order to see both the cold worked layer and the underlying unaffected core, it was necessary to use a wider flat of about .10 inch compared with .01 inch used by Wood. Therefore, a much lower magnification was obtained.

The bars, after running for a predetermined number of cycles as indicated by the counter attached to the motor, were removed from the machine. Following this, they were plated with two to three thousandths inch of nickel to preserve the edge of the flat during polishing. Plating was done in a nickel sulfate-nickel chloride Watt's type solution.

Over this was sprayed a coating of clear acrylic resin to insulate all of the specimen except the flat. The flat was then sanded along one side using emery paper and polished through 3/0 emery paper. Next, the flat was electropolished in orthophosphoric acid and finally was etched in ammonium persulfate. This etchant reveals the slip bands thus making the fatigue damage evident. Some difficulty was encountered, as will be evident in the photomicrographs, due to the fact that both the electropolishing and etching tended to attack the copper-nickel interface at a somewhat accelerated rate. Metallographic examination was made on a Bausch & Lomb Research Metallograph, and photomicrographs were taken using standard techniques.

Figure 1. Constant strain amplitude fatigue machine, front view.

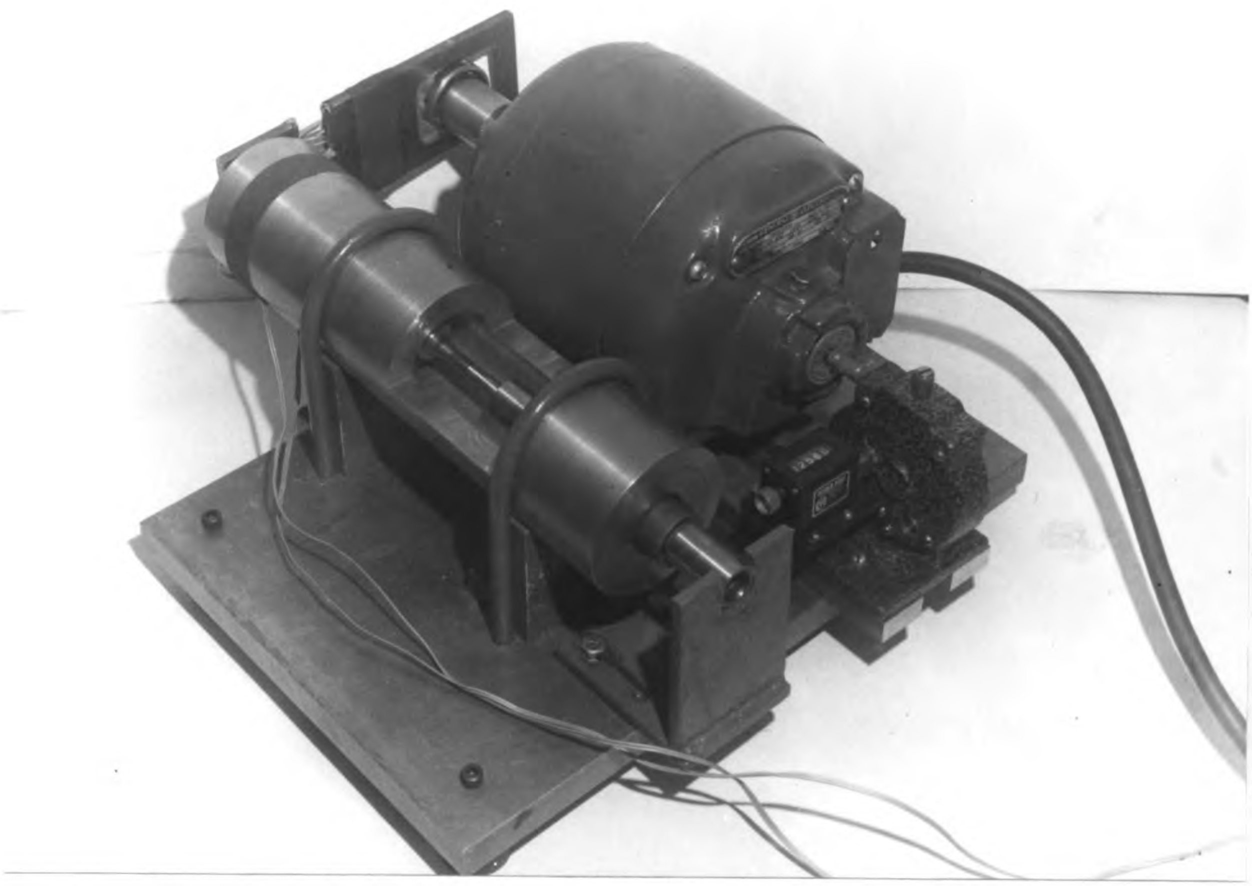

Figure 2. Constant strain amplitude fatigue machine, rear view.

# V. RESULTS AND DISCUSSION

The results of the metallographic inspection of strain cycled copper fatigue bars in the annealed and the peened conditions are summarized in Figures 3 through 28. In the unpeened condition, the following bars were taper sectioned and inspected: 640,000 cycles, 2,560,000 cycles, 10,920,000 cycles (failure). The bar which was run to failure was also stopped at a series of points throughout its life and the surface microphotographed. Shot peened bars were taper sectioned after the following numbers of cycles: 5000, 640,000, 2,560,000, 10,240,000, and 20,760,000. An additional bar was run to failure at 67,016,000 cycles which represents approximately a six-fold increase in life as a result of shot peening.

Microscopic examination of the surface of a polished specimen revealed slip band formation as expected. Figures 3 through 7 show an area on the surface at 100 magnification. Slip bands were visible after only 5000 cycles and as can be seen, as cycling continues the slip bands increase in width and darkness and new bands form. At a very early stage, the path of the final fracture is defined by slip bands. The exact point at which the fracture initiates is not evident but it is obvious that its path is almost entirely along slip bands. Figure 8 was taken at 500 magnification using oblique lighting. By this means the three dimensional nature of the slip bands is easily seen. Note the sharp notch-peak topography of the bands.

First observations of shot peened bars were made directly on the surface and little if any evidence of fatigue damage, such as slip bands, was visible, even after as many as 10,240,000 cycles of strain. This was originally interpreted to be due to the surface roughness arising from the shot peening. However, taper sectioning revealed a much more interesting and significant cause. A direct result of the peening is that slip band formation is inhibited in the surface layers. This is shown plainly in Figures 24 and 27. In these figures, etched up slip bands are visible in the interior of the bars but there is a region next to the surface where slip bands are few or non-existent. This is visible in Figures 24 and 27 at 100X. Compare this with Figure 20 which is of a polished bar and shows etched up slip bands right out to the surface. The effect is also evident in the photomicrographs taken at 500X.

The suppression of slip band formation in surface worked material could be attributed to three sources. First, the surface could be work hardened to a point where the strain which is being enforced in it remained purely elastic. Taking a value of the shear modulus of copper of six million and an approximate measured shear stress of 7000 psi, we get an elastic shear strain of .0012 inches per inch at the surface of the bar. The computed approximate applied shear strain is .0035 inches per inch so we see that some plastic deformation of the surface must be taking place. Secondly, this situation could be altered by the presence of the residual compressive stresses in the surface layers. That is, since the surface is already in compression, a greater tensile stress is required for it to reach its elastic limit in tension, so it can stand a greater load without yielding. However, in these tests we have used torsional loading which produces essentially pure shear in the surface. This shear can be thought of as a tension and a compression of equal magnitude at right angles to each other and at forty-five degrees to the shear stresses. The residual compression would then lower the stress in the tensile direction but would raise it in the compressive direction. Since the load is reversing, the net effect would be a cancellation. However, it is a known fact that with a constant stress range, the fatigue life is extended by a negative mean stress in push-pull fatigue tests. The third means by which surface working could produce the observed effect would arise from the dislocation distribution produced by the deformation of the surface. Recent experiments using transmission electron microscopy have shown that cold working of metals produces extremely complex, entangled dislocation arrays resulting from dislocation multiplication and interaction (39). It appears from the studies of annealed bars, that slip band formation requires the motion of dislocations over relatively large distances in a relatively narrow region. In the presence of dislocation entanglements, this would be very difficult due to the interactions between dislocations. It is possible then that in an annealed bar, plastic deformation at the surface is accomplished by the motion of a few dislocations over long distances but in the cold worked surface of the peened bars, it is by short movement of many dislocations. This would account for the absence of slip bands in the surface layer of the peened bars. It will be noted in the figures that the density of slip bands toward the center of the bars is approximately the same for peened and unpeened bars. This is in keeping with any of the above theories since they deal only with changes in the surface layer resulting from shot peening.

After 10 million cycles, fissures associated with etched up slip bands have formed in peened bars as can be seen in Figures 22 and 23. Also, after 20 million cycles, an intergranular crack was found and is shown in Figure 28. Although such microcracks were found at these points, failure did not occur in a peened bar until about 67 million cycles. When this is compared with a total life of an unpeened bar of 11 million cycles, it is quite evident that crack propagation is also inhibited by shot peening. This is possibly the way in which the compressive residual stress affects fatigue life. If the crack is being held closed by a residual compression, the stress at the tip due to the applied load will be lowered by a certain amount. The arguments applied above against this type of discussion for slip band formation are not so easily applied here. This is because slip is a result of shearing stresses while crack propagation is generally believed to be due to normal stresses.

It is also possible that a different mechanism of crack initiation than that proposed by Wood is operative in the fatigue of the shot peened bars. This is suggested by the intergranular crack which is seen in Figure 28. However, fissures associated with slip bands are seen, as already mentioned, in Figures 22 and 23. So, it is not certain whether the fatigue crack is initiated by slip and fissure formation at a weak point in the deformed layer or some other mechanism is operative. In any case, it is obvious that further work is needed to separate the effects of cold work and residual stress on the fatigue behavior of surface worked material.

The etched up markings which were visible after 5000 cycles of strain are shown in Figures 9 and 10. There is some doubt as to whether or not these are slip markings.

They appear to be crystallographic in nature, lying along three distinct directions. However, they do not lie in the direction of maximum shear as one would expect slip markings to do. Furthermore, there seems to be some difference in appearance as can be seen in Figure 23 where both these markings and slip bands are present. It is possible that these markings are the result of some type of strain induced gathering of lattice imperfections. This could account in part for the observed softening of cold worked material under alternating strain.

There is also an apparent difference in appearance of the etched up slip bands in peened and unpeened specimens. In peened bars, small gaps appear in the bands as if a pore had formed in the band. As the numbers of cycles increases, the length of the gaps increases. These gaps are seen in unpeened bars only near the surface when fissure formation has occurred and are more prevalent in peened bars near the center than near the surface. These gaps or pores could form by the coalescence of vacancies which is possible at room temperature. Since the core must be under a slight residual tension to balance the compressive surface, these points of high vacancy concentration could be opened up into pores. This could be checked by running a bar at low temperature so that diffusion is greatly slowed and then seeing if the gaps are present.

The final point to be discussed is the manner in which the shear stress varied with the number of cycles in peened and unpeened bars. Figure 29 shows this variation schematically. The annealed and polished bars behaved as expected from previous work. The peak stress for the first cycle was about 2500 psi. This rose to approximately 4000 psi after 100 cycles, 5000 psi after 1000 cycles and 7000 psi after 10,000 cycles. After remaining constant at this level for several hundred thousand cycles, the stress gradually faded as fatigue damage progressed until fracture occurred. Peened bars on the other hand, started out at about 7000 psi, rose slightly to perhaps 7200 psi and then dropped to a value near 6000 psi after a few hundred thousand cycles. This value was then maintained for millions of cycles until it finally dropped as a visible crack propagated to fracture. The gradual fading of the stress in the polished bars as opposed to the constant stress in the peened bars is probably due to the gradual spread of fatigue damage in the surface and the formation of fissures in unpeened specimens. Since these occurrances are inhibited by peening, there is no gradual fading. The initial rise may be due to work hardening of the core material in the peened bars.

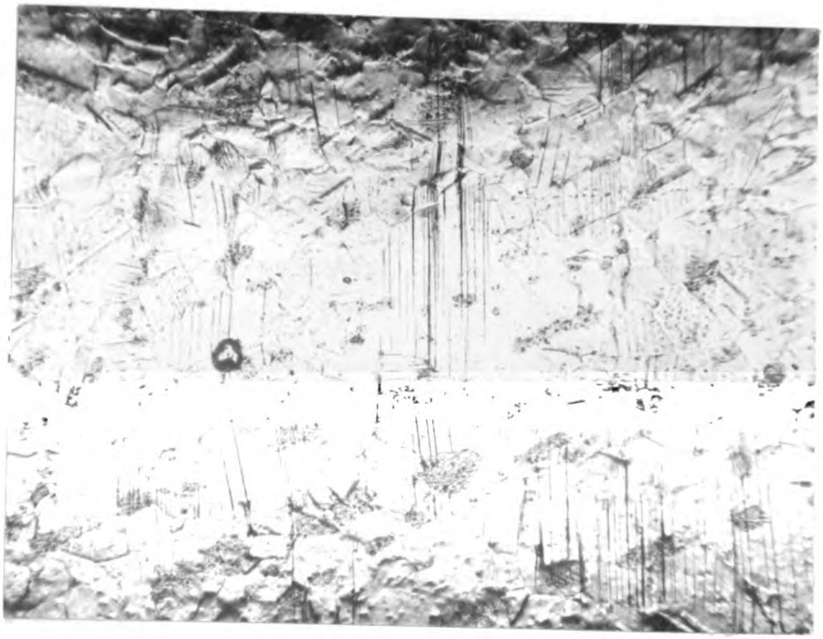

Figure 3. Surface of annealed and polished bar after 80,000 cycles. 100x.

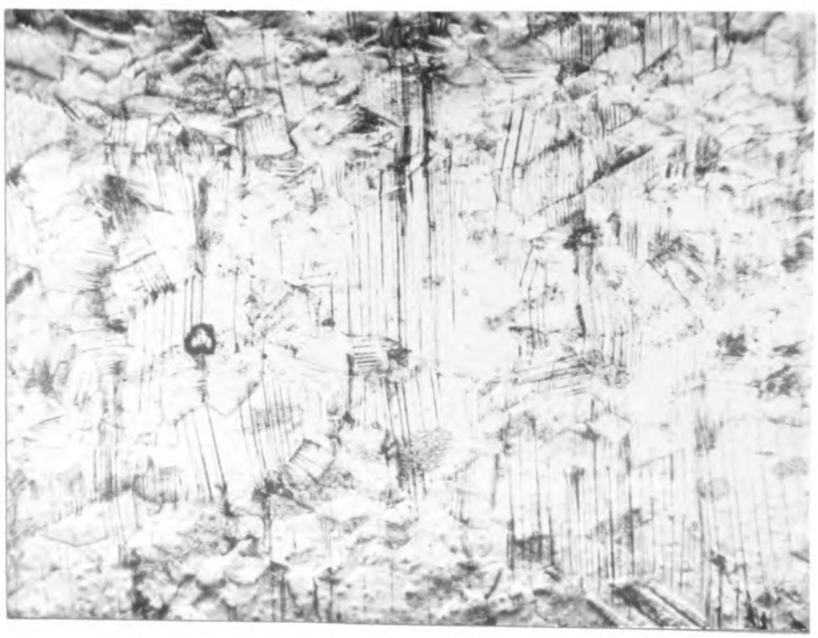

Figure 4. Surface of annealed and polished bar after 640,000 cycles. 100x.

Figure 5. Surface of annealed and polished bar after 2,560,000 cycles. 100x.

Figure 6. Surface of annealed and polished bar after 5,120,000 cycles. 100x.

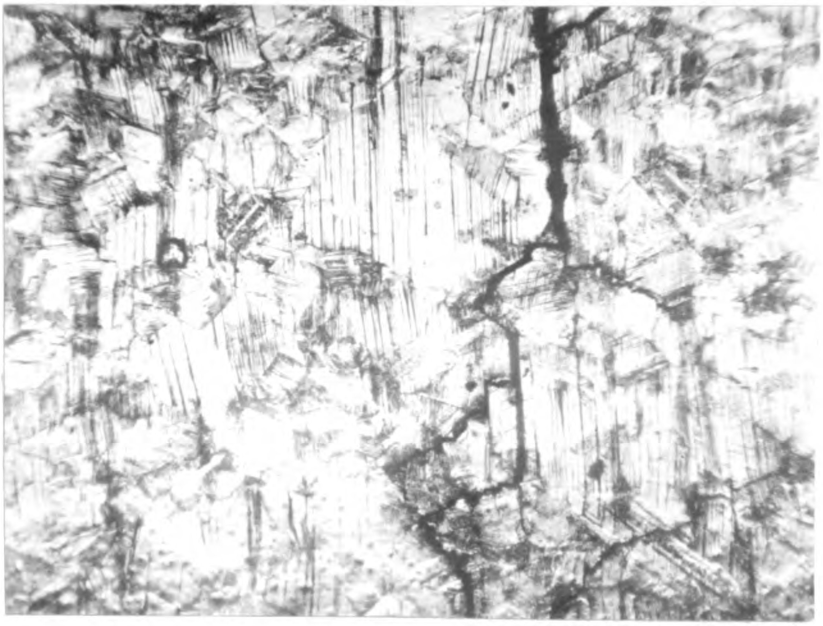

Figure 7. Surface of annealed and polished bar after 10,240,000 cycles. 100x.

Figure 8. Surface of annealed and polished bar after 320,000 cycles. 500x. Oblique light.

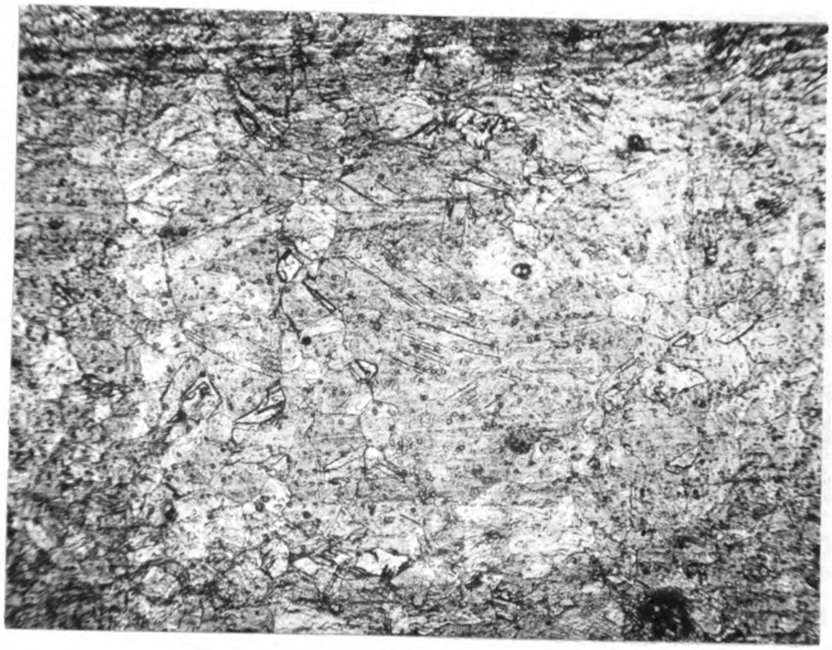

Figure 9. Taper section of peened bar after 5000 cycles. 100x.

Figure 10. Taper section of peened bar after 5000 cycles. 500x.

Figure 11. Taper section of peened bar after 640,000 cycles. Etched up markings near surface. 500x.

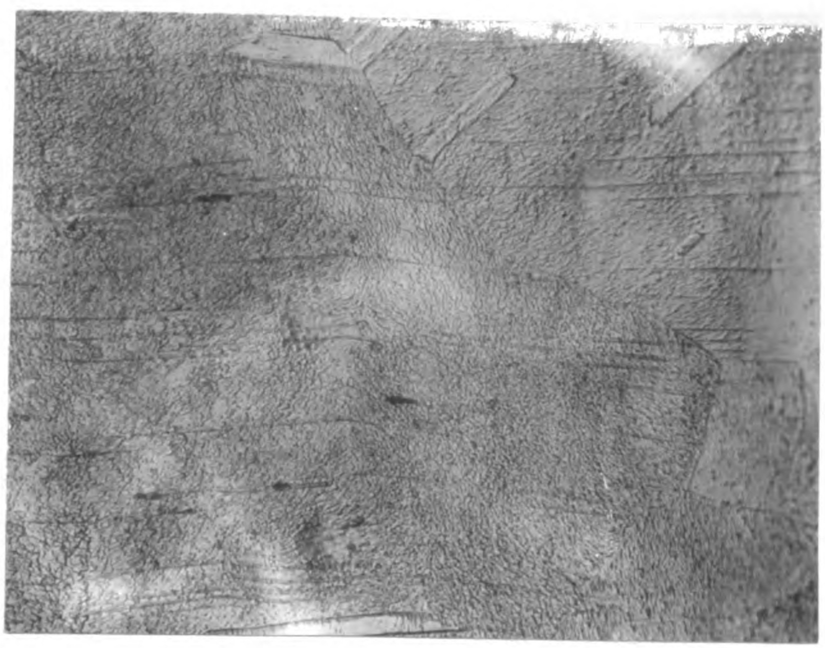

Figure 12. Taper section of peened bar after 640,000 cycles. Etched up slip bands near the center of the taper section. 500x.

Figure 13. Taper section of polished bar after 640,000 cycles. Etched up slip bands near surface. 500x.

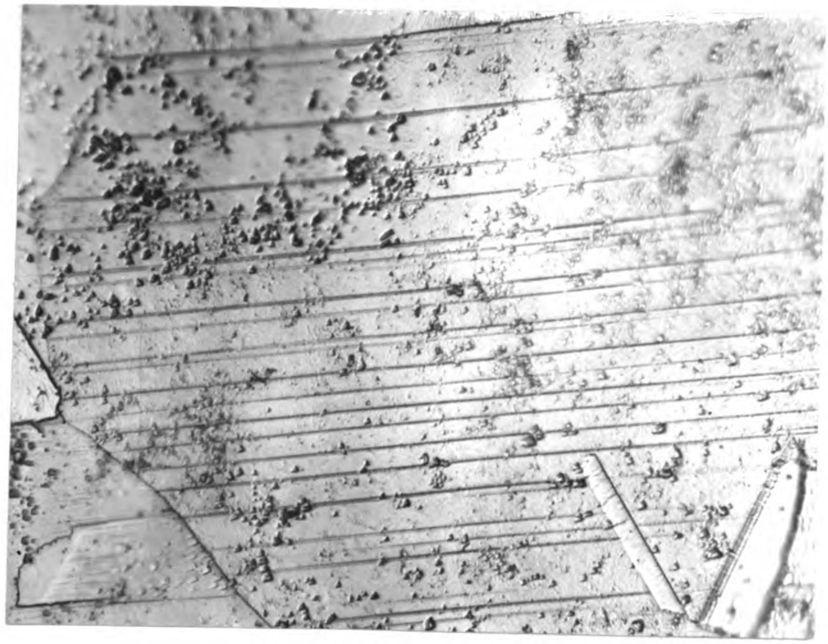

Figure 14. Taper section of polished bar after 640,000 cycles. Etched up slip bands at center of taper section. 500x.

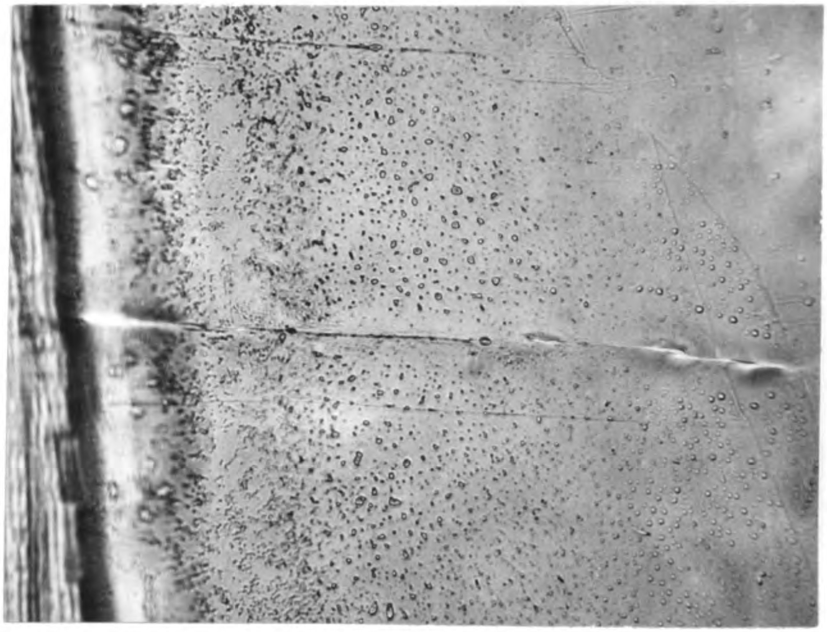

Figure 15. Taper section of polished bar after 640,000 cycles. Fissures visible at the surface before etching. 500x.

Figure 16. Taper section of peened bar after 2,560,000 cycles. Etched up slip bands near but not at the surface.

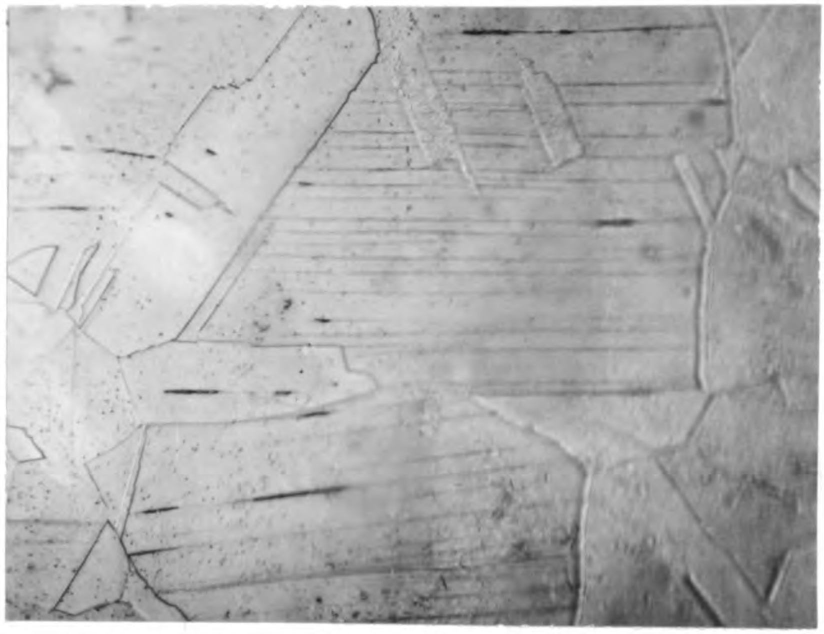

Figure 17. Taper section of peened bar after 2,560,000 cycles. Etched up slip bands in the center of the taper section. 500x.

Figure 18. Taper section of polished bar after 2,560,000 cycles. Fissures and etched up slip bands at the surface. 500x.

Figure 19. Taper section of polished bar after 2,560,000 cycles. Etched up slip bands near the center of the taper section.

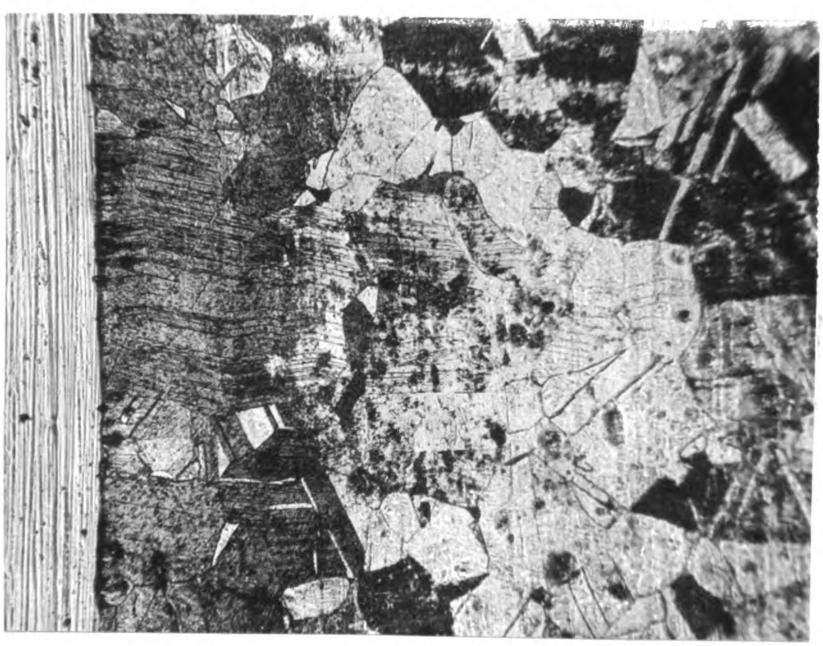

Figure 20. Taper section of polished bar after 2,560,000 cycles. Etched up slip bands at the surface. 100x.

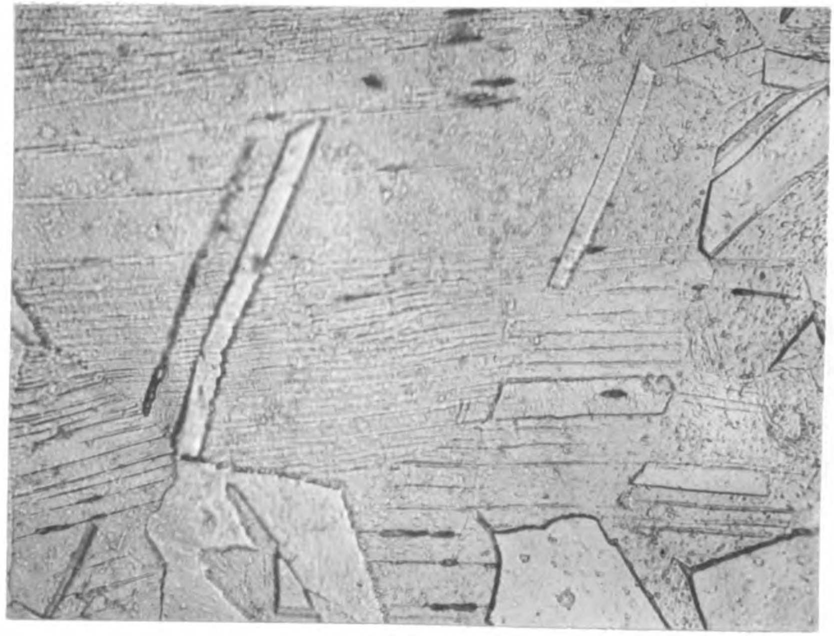

Figure 21. Taper section of peened bar after 10,240,000 cycles. Etched up slip bands near the center of the taper section. 500x.

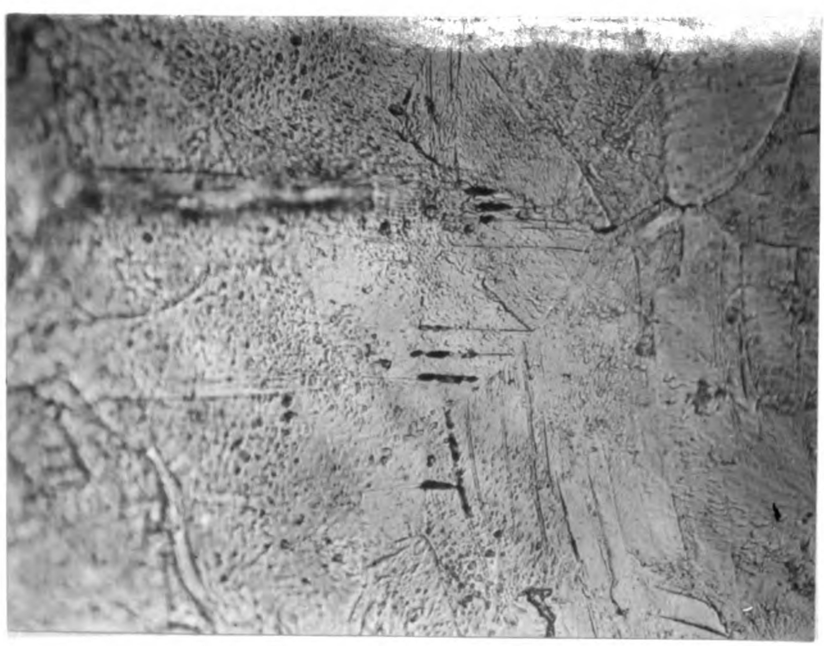

Figure 22. Taper section of peened bar after 10,240,000 cycles. Fissure and etched up slip at the surface. 500x.

Figure 23. Taper section of peened bar after 10,240,000 cycles. Fissures and etched up slip markings near the surface. 500x.

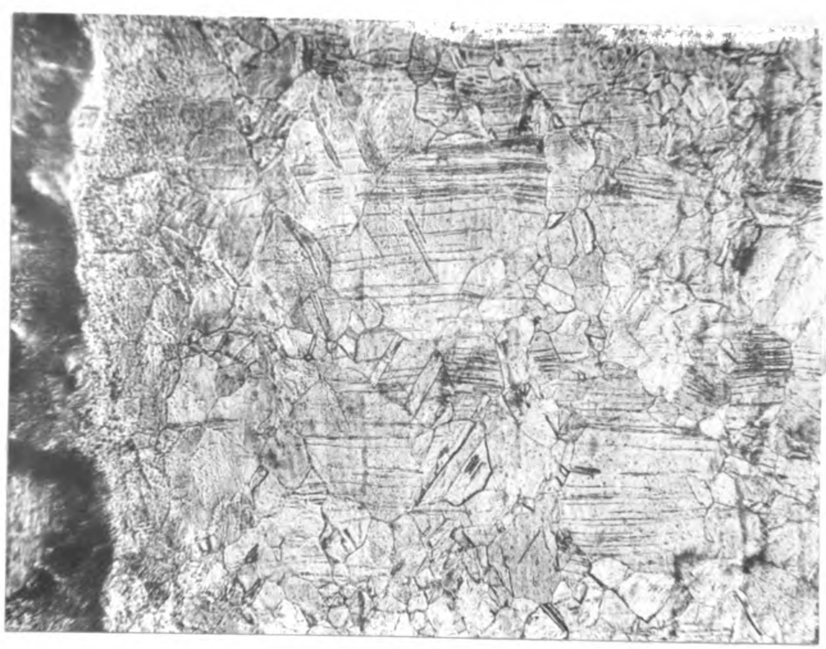

Figure 24. Taper section of peened bar after 10,240,000 cycles showing very little slip in the surface layer. 100x.

Figure 25. Taper section of polished bar after fracture at 10,920,000 cycles. Etched up slip bands near the fracture. 500x.

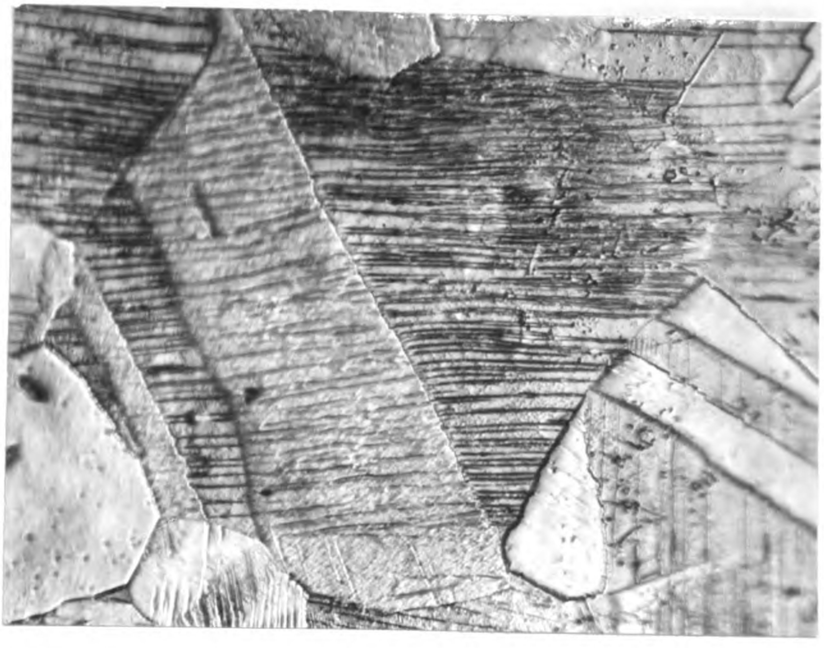

Figure 26. Taper section of peened bar after 20,000,000 cycles. Etched up slip bands in the center of the taper section. 500x.

Figure 27. Taper section of peened bar after 20,000,000 cycles showing reduced number of slip bands at the surface. 100x.

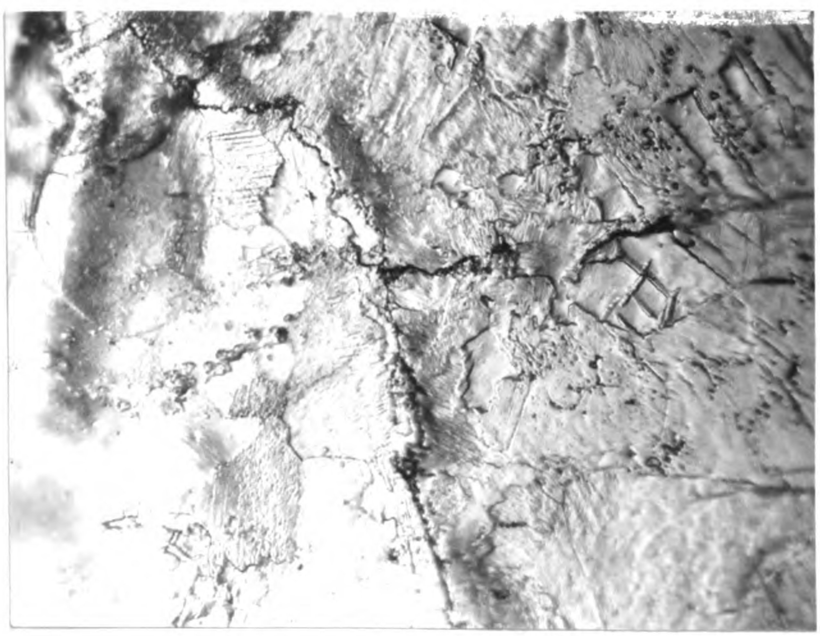

Figure 28. Intercrystalline crack visible in Figure 27 at 500x.

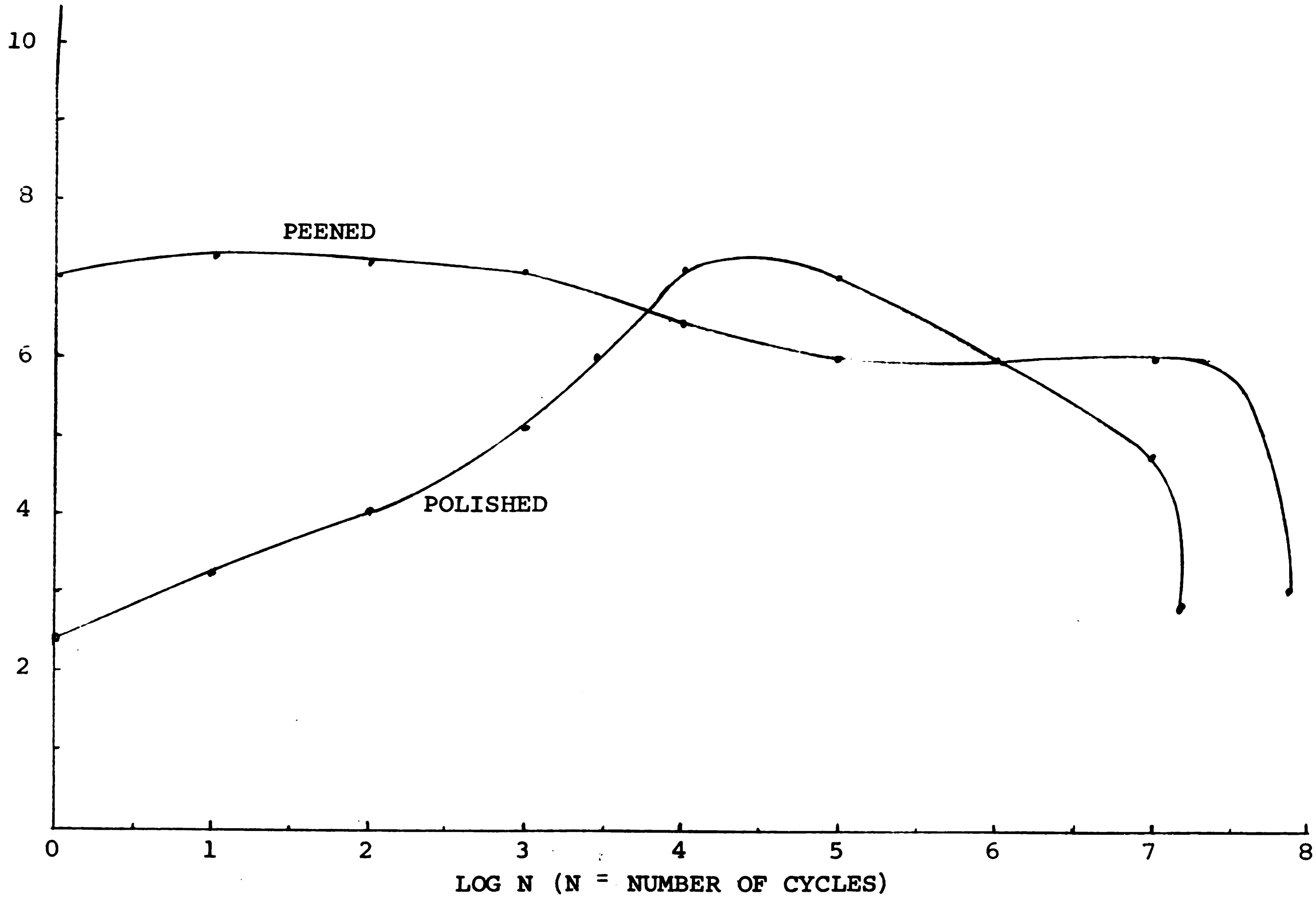


Figure 29. Peak stress vs. log of number of cycles of strain for polished and peened bars (schematic).

# VI. CONCLUSIONS

1. Significant differences in the fatigue process on the microscopic level were noted between polished bars and shot peened bars.

2. The fatigue life of shot peened bars is extended over that of polished bars by either or both of the following means:

   A. Inhibition of slip band formation in the surface layer.

   B. Reduced rate of crack propagation.

   A marked decrease in slip band density near the surface of shot peened bars was found upon metallographic examination. The number of cycles between observation of microcracks and failure in peened bars was greater than the total life of unpeened bars.

3. The nucleation of fatigue cracks at fissures (intrusions) in slip bands, as proposed by Wood for annealed and polished bars, may not be the operative mechanism in shot peened bars. Both fissures and intercrystalline cracks were found in peened bars while only fissures were found in polished bars.

4. The present work does not clearly establish whether the alteration in the fatigue process in shot peened bars results from compressive residual stresses or cold work.

# VII. RECOMMENDATIONS

The present work indicates that significant differences are produced in the slip band structure by shot peening annealed copper fatigue specimens. However, much work is needed to clarify the mechanism by which these differences are obtained. Some avenues for consideration for future work along this line are:

1. Examination of slip interference by controlled arrays of dislocations near the surface of single crystals in the absence of residual stresses, and study of the effects on the fatigue life. Such controlled arrays could be produced by a polygonization anneal after surface cold work.
2. Examination of slip interference and effect on fatigue life by various other dislocation obstacles near the specimen surface. Specimens could be dispersion hardened near the surface by diffusion alloying of the surface layer.
3. Development of an etch pit technique for the study of dislocation processes during the fatigue process in both peened and annealed bars.
4. Study of the change in degree of cold work and residual stress in the surface layer as a result of cyclic

straining by X-ray and microhardness methods. Cold work affects the breadth of the Debye rings in X-ray patterns while residual stress affects their location.

Certain recommendations of a mechanical nature are also in order:

A. Use of a surface working technique which leaves the surface in a smooth condition. This would allow observation of surface as well as subsurface effects. Surface rolling or burnishing would be suitable in this respect.

B. Redesign of the specimen ends and the specimen holder shafts to allow easier installation of specimens with less chance of straining during installation and more positive application of the loads. Proper design would also allow the use of one of the holder shafts as a weigh bar which would be better than using the lever arm as presently done.

C. Use of a slightly larger strain amplitude to accelerate the tests and give larger torques which could be measured more accurately.

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