STRAIN FIELD NEAR THE SURFACE
DUE TO SURFACE TRACTION

by

GUDMUNDUR AGUSTSSON
Odense Teknikum
(1971)

Submitted in Partial Fulfillment
of the Requirements for the
Degree of Master of Science
at the
MASSACHUSETTS INSTITUTE OF TECHNOLOGY
March, 1974

Signature redacted

Signature redacted

Certified by.................................................................
Thesis Supervisor

Signature redacted

Accepted by.................................................................
Chairman, Departmental Committee
on Graduate Students
STRAIN FIELD NEAR THE SURFACE DUE TO SURFACE TRACTION

by

GUDMUNDUR AGUSTSSON

Submitted to the Department of Mechanical Engineering on March 5, 1974 in partial fulfillment of the requirements for the degree of Master of Science.

ABSTRACT

A test where plasticine is deformed under pure shear shows that the deflection of grain boundaries can be used to determine plastic strain. Surface traction is created when two members slide against each other. The surface traction deforms the surface layer. The deformation is determined by observing the deflection of the grain boundaries. The effective plastic strain is greatest at the surface and decreases with increasing distance from the surface. The magnitude of deformation and thickness of the deformed layer depends on the member's metallurgical and mechanical properties. In the case of AISI 1020 steel, the thickness of the deformed layer was Ca. 40 μm and the effective plastic strain at the surface was as high as 16.5. In copper the thickness of the deformed layer was Ca. 95 μm and the effective plastic strain at the surface apparently of the order $1 \times 10^2$. 

Thesis Supervisor: Nam P. Suh

Title: Associate Professor of Mechanical Engineering
ACKNOWLEDGMENTS

It is a pleasure to acknowledge here my appreciation for the help that I have received from my supervisor, Professor Nam P. Suh.

I would like to thank Mr. Said Jahanmir for his valuable advice. Also, I am grateful to Mr. Christopher Rotz for editing the manuscript of this thesis.

Appreciation is expressed to the personnel of the Materials Processing Laboratory for their assistance and good companionship during the course of this project. Thanks to Maureen Rush for the excellent typing of this thesis.
# TABLE OF CONTENTS

<table>
<thead>
<tr>
<th>Section</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>ABSTRACT</td>
<td>2</td>
</tr>
<tr>
<td>ACKNOWLEDGMENTS</td>
<td>3</td>
</tr>
<tr>
<td>CHAPTER I - INTRODUCTION</td>
<td>6</td>
</tr>
<tr>
<td>CHAPTER II - QUANTITATIVE DETERMINATION OF SHEAR STRAIN.</td>
<td>7</td>
</tr>
<tr>
<td>2.1 Introduction to the Quantitative Determination of a Strain Field Due to Pure Shear</td>
<td>7</td>
</tr>
<tr>
<td>2.2 Mathematical Derivation of the Effective Plastic Strain by Observing the Deflection of the Grain Boundaries</td>
<td>7</td>
</tr>
<tr>
<td>2.3 Experimental Testing</td>
<td>12</td>
</tr>
<tr>
<td>2.4 Experimental Results</td>
<td>15</td>
</tr>
<tr>
<td>CHAPTER III - EFFECTS OF THE SURFACE TRACTION ON SURFACE LAYER</td>
<td>18</td>
</tr>
<tr>
<td>3.1 Surface Traction and Strain Creation</td>
<td>18</td>
</tr>
<tr>
<td>3.2 Void Nucleation in the Sub-Surface Layer</td>
<td>19</td>
</tr>
<tr>
<td>3.3 Hole Growth in the Sub-Surface Layer</td>
<td>24</td>
</tr>
<tr>
<td>CHAPTER IV - DETERMINATION OF SHEAR STRAIN IN METAL</td>
<td>27</td>
</tr>
<tr>
<td>4.1 Introduction to the Linear Intercept Method</td>
<td>27</td>
</tr>
<tr>
<td>4.2 Mathematical Derivation of the Linear Intercept Method</td>
<td>27</td>
</tr>
<tr>
<td>CHAPTER V - EXPERIMENTAL PROCEDURE</td>
<td>31</td>
</tr>
<tr>
<td>5.1 Specimen Preparation</td>
<td>31</td>
</tr>
<tr>
<td>5.2 Testing Equipment</td>
<td>33</td>
</tr>
<tr>
<td>5.3 Specimen Preparation for Microscopic Observations</td>
<td>34</td>
</tr>
<tr>
<td>5.4 Measurement Equipment and Method</td>
<td>37</td>
</tr>
<tr>
<td>CHAPTER VI - EXPERIMENTAL RESULT</td>
<td>39</td>
</tr>
<tr>
<td>6.1 Determination of Strain Field</td>
<td>39</td>
</tr>
<tr>
<td>6.2 Void Formation</td>
<td>41</td>
</tr>
<tr>
<td>6.3 Crack Formation</td>
<td>43</td>
</tr>
<tr>
<td>CHAPTER VII - DISCUSSION</td>
<td>46</td>
</tr>
</tbody>
</table>
Table of Contents (continued)

<table>
<thead>
<tr>
<th>Section</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>CHAPTER VIII - SUMMARY AND CONCLUSION</td>
<td>47</td>
</tr>
<tr>
<td>REFERENCES</td>
<td>48</td>
</tr>
<tr>
<td>LIST OF TABLES</td>
<td>50</td>
</tr>
<tr>
<td>LIST OF FIGURES</td>
<td>51</td>
</tr>
</tbody>
</table>
CHAPTER I

INTRODUCTION

The conventional theories of sliding wear are basically only described by interaction of the contacting surfaces. The material removal is said to take place by adhesion or abrasion. This has resulted in concentrated investigation of the worn surface. When material right under the surface is observed new aspects of wear are introduced. First of all, what is noticed when the material next to the surface is studied is that the surface layer is heavily deformed. This high degree of deformation results in several things. Where second phase hard particles are present in the deformed layer, voids are created around them. These voids become holes whose growth depends on the magnitude of deformation. The holes can join together and form wear sheets.

In this project a model to determine the plastic strain in metal resulting from pure shear is tested. The model is based on the fact that the alteration in grain thickness is proportional to the deformation.

Strain field resulting from surface traction is calculated and investigated.
CHAPTER II

QUANTITATIVE DETERMINATION OF SHEAR STRAIN

2.1 Introduction to the Quantitative Determination of a Strain Field Due to Pure Shear

When two materials slide against each other under pressure without any lubricant, substantial plastic deformation takes place at the surface. This deformation is considered to be a shear deformation.

Dautzenberg [1] has introduced methods to determine the effective deformation by measuring the geometrical alteration of the metal grains.

This method is based on the fact that when an ideal spherical grain is deformed under pure shear it becomes an ellipsoid. Therefore, when a cut is made in the material parallel to the sliding direction and perpendicular to the sliding surface the undeformed ideal spherical grains will appear as circles and the deformed grains as ellipses. The effective plastic strain can then be determined by observing the geometrical deformation of the grains when they change from spheres to ellipsoids.

2.2 Mathematical Derivation of the Effective Plastic Strain by Observing the Deflection of the Grain Boundaries

The incremental effective plastic strain, \( d \bar{\varepsilon} \), is defined as [2]:
\[
\frac{3}{4} \left( \frac{d\varepsilon_x}{\sigma} \right)^2 = \frac{1}{2} \left( \frac{d\varepsilon_x}{\sigma} - \frac{d\varepsilon_y}{\sigma} \right)^2 + \left( \frac{d\varepsilon_y}{\sigma} - \frac{d\varepsilon_z}{\sigma} \right)^2 + \left( \frac{d\varepsilon_z}{\sigma} - \frac{d\varepsilon_x}{\sigma} \right)^2 \\
+ \frac{3}{4} \left[ d\gamma_{yx}^2 + d\gamma_{zy}^2 + d\gamma_{xz}^2 \right]
\]  

(1)

and the effective stress, \( \bar{\sigma} \), is defined as [2]:

\[
2\bar{\sigma}^2 = (\sigma_x - \sigma_y)^2 + (\sigma_y - \sigma_z)^2 + (\sigma_z - \sigma_x)^2 \\
+ 6\tau_{yx}^2 + 6\tau_{zy}^2 + 6\tau_{xz}^2 = 6K^2
\]

(2)

Under pure shearing and plain strain:

\[
d\varepsilon_x = d\varepsilon_y = d\varepsilon_z = d\gamma_{yz} = d\gamma_{yx} = 0
\]

and

\[
d\gamma_{xy} \neq 0.
\]

The Levy von Mises equations are:

\[
\frac{d\varepsilon_x}{\sigma} = \frac{d\varepsilon}{\sigma} \left( \sigma_x - \frac{\sigma_y + \sigma_z}{2} \right) ; \quad d\gamma_{yx} = 3 \frac{d\varepsilon}{\sigma} \tau_{yx}
\]

\[
\frac{d\varepsilon_y}{\sigma} = \frac{d\varepsilon}{\sigma} \left( \sigma_y - \frac{\sigma_z + \sigma_x}{2} \right) ; \quad d\gamma_{zy} = 3 \frac{d\varepsilon}{\sigma} \tau_{zy}
\]

\[
\frac{d\varepsilon_z}{\sigma} = \frac{d\varepsilon}{\sigma} \left( \sigma_z - \frac{\sigma_x + \sigma_y}{2} \right) ; \quad d\gamma_{xz} = 3 \frac{d\varepsilon}{\sigma} \tau_{xz}.
\]

(4)

Equations (3) and (4) give:
\[ \tau_{zy} = \tau_{xz} = 0 \quad \text{and} \]
\[ \sigma_x = \frac{\sigma_y + \sigma_z}{2} = 0 \]
\[ \sigma_y = \frac{\sigma_z + \sigma_x}{2} = 0 \quad \sigma_x = \sigma_y = \sigma_z \quad (5) \]
\[ \sigma_z = \frac{\sigma_x + \sigma_y}{2} = 0 \]

Equation (5) into (2) gives
\[ 2 \frac{\sigma^2}{\sigma} = 6\tau^2_{yx} = 6k^2 \]
\[ \sigma = \tau_{yx} \sqrt{3} \quad (6) \]

Equations (1) and (3) give:
\[ \frac{9}{4}(d\overline{\epsilon})^2 = \frac{3}{4} d\gamma_{xy}^2 \]
\[ d\overline{\epsilon} = \frac{d\gamma_{xy}}{\sqrt{3}} \]
\[ d\overline{\epsilon} = \frac{1}{\sqrt{3}} \frac{dl_x}{\gamma_y} \]
\[ \overline{\epsilon} = \frac{1}{\sqrt{3}} \int_{l_y}^{l_x} dl = \frac{1}{\sqrt{3}} \frac{l_x}{\gamma_y} \quad . \quad (7) \]
In Fig. 1 \( \lambda_y = OA \) and \( \lambda_x = AB \)

so \( \tan \gamma = \frac{\lambda_x}{\lambda_y} \) and

\[
\varepsilon = \frac{\tan \gamma}{\sqrt{3}}
\]  

(8)

where \( \gamma \) is the angle of shear.

To determine \( \varepsilon \) a relationship between \( \tan \gamma \) and the decrease in grain thickness is needed.

To obtain this let us observe Fig. 1. The circle has its centrum at the point \((0, \frac{1}{2}D)\) and radius is \( \frac{1}{2}D \), or

\[
\frac{D}{2} = \sqrt{x^2 + (y - \frac{1}{2}D)^2}
\]

(9)

After pure shear over the angle \( \gamma \) the point \((x, y)\) of the circle goes over to the point \((x^1, y^1)\) of the ellipse and the relationships between these coordinates are:

\[
x^1 = x + y^1 \tan \gamma
\]

\[
y^1 = y
\]

(10)

Equation (10) into (9) gives

\[
(x^1 - y^1 \tan \gamma)^2 + (y^1 - \frac{1}{2}D)^2 = \frac{D^2}{4}
\]
or

\[ x'^2 + y'^2(1 + \tan^2 \gamma) - 2x'y' \tan \gamma - y'D = 0 \quad (11) \]

The length of the chord \( C \) can be determined by finding the \( y' \) coordinates of the intersections which the line \( x = \frac{1}{2} D \tan \gamma \) makes with the ellipses boundary, and then subtracting the lower \( y' \) value from the bigger \( y' \) value.

\( x' = \frac{1}{2} D \tan \gamma \) into (11) gives:

\[ y'^2 - y'D + \frac{1}{4} D^2 \left( \frac{\tan^2 \gamma}{1 - \tan^2 \gamma} \right) = 0 \]

\[ y'^2 - y'D + \frac{1}{4} D^2 \sin^2 \gamma = 0 \quad (12) \]

or

\[ y' = \frac{D}{2} \pm \frac{D}{2} \sqrt{1 - \sin^2 \gamma} \]

\[ y' = \frac{D}{2} \pm \frac{D}{2} \cos \gamma \]

so

\[ C = \frac{D}{2} + \frac{D}{2} \cos \gamma - \frac{D}{2} + \frac{D}{2} \cos \gamma \]

\[ C = D \cos \gamma \quad . \quad (13) \]

Now

\[ \tan^2 \gamma = \frac{\sin^2 \gamma}{\cos^2 \gamma} = \frac{1 - \cos^2 \gamma}{\cos^2 \gamma} = \frac{1}{\cos^2 \gamma} - 1 \quad . \]
Now
\[
\tan^2 \gamma = \frac{\sin^2 \gamma}{\cos^2 \gamma} = \frac{1 - \cos^2 \gamma}{\cos^2 \gamma} = \frac{1}{\cos^2 \gamma} - 1
\]

Equation (13) gives
\[
\cos^2 \gamma = \frac{c^2}{d^2}
\]

so
\[
\tan \gamma = \sqrt{\frac{d^2}{c^2} - 1}
\]

and Eq. (8) becomes
\[
\bar{\varepsilon} = \left[ \frac{\frac{d^2}{c^2} - 1}{3} \right]^{1/2}.
\]

So in the ideal case the effective plastic strain can be determined by measuring the diameter before deformation and the C value after deformation.

2.3 Experimental Testing

An experiment was done to test the above method of determining the plastic strain by measuring the decrease in grain thickness.

a. Equipment:

A test instrument was made to impose a pure shear deformation on a specimen made of plasticine. Figure 2 shows a sketch of the instrument. It consists of a baseplate and a
rectangular bar sliding in a groove in the baseplate. The bar is supported by two bar holders and can be moved by screwing a bolt. The movements are measured by a micrometer. A specimen holder is connected to the bar and moves relative to the baseplate as the bar is moved. Another specimen holder can be fastened to the baseplate at different places depending on the size of the specimen. Finally there is a movable reference axis.

b. Procedure:

Two plasticine specimens were made, each 6 inches wide, 7 inches long and 1.25 inches thick. Circular marks were put on one surface of one of the plasticine pieces by placing a circular rod with a concave end in a "tape-o-matic" drill press. As the position of the drill press table was digitally controlled, even spacing of the marks was achieved within 1/1000 inch.

When the marks had been made on the piece it was placed in the test device with the marked side up. Then everything was lined up: the scale was adjusted so it was perpendicular to the movable bar and level with the upper surface of the plasticine sheet. The holders were fixed so they were parallel to the bar. The plasticine sheet was then placed in the holders so the rows of circular marks were parallel to the bar. See Fig. 8. The whole set-up was then placed under a camera. Two pictures were taken: one paperprint (Fig. 5) for later reference and one slide for measurement.
The upper surface of the sheet was then lightly covered with powder and the second sheet placed upon the first one without moving the test device. This gave the specimen a 2.5 inch overall thickness which insured plane strain deformation. Shear deformation was then applied by screwing the bolt.

After deformation the upper sheet and the remaining powder were removed from the upper surface of the lower sheet. Then a slide and photoprint were taken of the deformed circles. Fig. 6. Between pictures great care was taken to insure no vertical displacement between the test instrument and the camera.

C. Measurements:

Measurements were made by placing the slide between two glass pieces. Then the glasses with the slide were placed in an optical comparator with 10x magnification so that relatively exact measurements could be made. Fig. 3.

First, the diameter of the undeformed circles was measured by taking 7 arbitrary circles and measuring the longitudinal and the transverse diameters of each circle. The average of these two measurements was then assumed to be the diameter of that circle. The average diameter was then determined by taking the average of the 7 diameters.

Then the slide picture of the deformed stage was placed in the optical comparator in the same manner and adjusted so the reference axis on the film coincided with the axis on the screen. To be able to locate the intersection of
the screen axes in the centrum of the ellipse a system of equally spaced parallel lines was placed parallel to each of the axes near their intersection. The centrum of the ellipse was then found by positioning the axes of the scope parallel to the major and minor axes of the ellipse and counting the number of lines on each side of each axis from the centrum to the edges of the ellipse. Fig. 4. Once the centrum of the ellipse was found, the "C" value, which is as stated earlier the length of the linear intersection which goes through the ellipse centrum parallel to the y-axis, Fig. 1, was easily measured by moving the optical comparator table. This was done by positioning the intersection of the screen axes on one edge of the ellipse and then moving the table so the intersection fell on the opposite edge of the ellipse. The distance traversed by the table was then determined by the micro-screw which moves the table.

2.4 Experimental Results

The average diameter was calculated to be .2544 inch. The biggest variation from the average was .0016 inch or 0.63%.

Figure 6 shows the plasticine in the deformed stage. It can be observed from the picture that pure shear deformation has only taken place inside the marked parallelogram. Therefore, only the five ellipses inside the parallelogram were chosen for the measurement of the "C" value.

The "applied" angle of shear was:
\[ \gamma_{\text{APPL}} = 15.08^\circ \]

and Eq. (8) gives

\[ \gamma_{\text{APPL}} = 0.153 \]

With a careful look at Fig. 6 one can see that a small amount of rotation has taken place within the parallelogram. This rotation was measured to be 0.57°. Thus the "true" angle of shear inside the parallelogram was:

\[ \gamma_{\text{TRU}} = \gamma_{\text{APPL}} - \gamma_{\text{ROT}} \]

\[ \gamma_{\text{TRU}} = 15.08^\circ - 0.57^\circ = 14.51^\circ \]

so

\[ \gamma_{\text{TRU}} = 0.1495 \]

The average "C" value was 0.2471 inch and the biggest variation from the average was 0.0027 inch or 1.8%.

With \( D = 0.2544 \) inch and \( C = 0.2471 \) inch Eq. (14) gives

\[ \bar{\varepsilon} = 0.1414 \]

which differ 7.6% from \( \bar{\varepsilon}_{\text{APPL}} \) and 5.4% from \( \bar{\varepsilon}_{\text{TRU}} \) which \( \bar{\varepsilon} \) is less.

The most probable reason for this difference is the lack of rigidity of the test device and most likely a small relative displacement between the specimen and specimen holder.
This experiment shows that the previous mathematical derivations are valid and that in the ideal circular case the effective plastic strain in pure shear can be determined with good accuracy.
CHAPTER III
EFFECTS OF THE SURFACE TRACTION ON SURFACE LAYER

3.1 **Surface Traction and Strain Creation**

Surface traction between two metal objects which slide on each other is conventionally described by the adhesion theory of friction where the friction is created by adhesion between asperities [3]. Recent investigation of the wear phenomenon indicates that the surface traction between nonlubricated sliding surfaces is much higher than can be explained by the adhesion theory.

Suh [4] observed a heavily deformed sub-surface layer in a wear specimen which cannot be described by adhesion of asperities only. This big surface traction is explained by a "plowing" mechanism which exists when asperities of the harder body sink into the softer one, resulting in much higher friction force.

According to "The Delamination Theory of Wear," Suh [5], deformation of the sub-surface layer causes cracks to be formed under the surface. These cracks lie parallel to the surface, and where they reach a certain critical length they extend to the surface and a wear sheet is formed.

Jahanmir, et al. [6] has observed that the upper surface of a wear sheet prior to removal from the material has a relatively smooth surface with parallel furrows in the sliding direction. This indicates that once the asperities of the softer material are worn away, which can happen very quickly, the surface traction is a combination of adhesion and the
"plowing" mechanism. When the surface layer on the specimen is relatively much softer than the slider's surface layer, adhesion and "plowing" will act instantaneously.

This high surface traction causes high shear deformation of the sub-surface layer. The extension of this deformed layer into the material depends on the composition of the material involved.

The shearing apparently does not take place all at once, but is a result of a repeated surface traction at each surface point caused by passing asperities. Each time asperities pass a point on the surface a certain amount of shear force is transmitted from the surface layer to the sub-surface layer, resulting in a small amount of deformation. This repeated incremental deformation accumulates in the deformed layer, causing it to shear more and more. This shearing deformation will then continue until some kind of steady-state condition is established. That happens when the work due to the friction force is dissipated by deformation of new material as the surface layer wears off, additional deformation of the already deformed material and heat generation. The deformation is therefore biggest at the surface and decreases with increasing distance from the surface.

3.2 Void Nucleation in the Sub-Surface Layer

It has been well known for a long time that second phase particles play a fundamental role in the ductile fracture of metal. The reason for this is that when the metal is strained voids are primarily formed around the large second phase
particles. The mechanism of the void formation has been investigated by a number of people but has not yet been fully understood.

As strain due to surface traction is very high in the deformed layer, one could expect that void would be formed around hard particles somewhere within that layer. This has been observed by Jahanmir [7].

In the chapter about surface traction and shear deformation, it was explained how the shear force due to the surface traction was transmitted from the surface layer down to the sub-surface layer. This shear stress acting under the surface layer will generate dislocations. The dislocation density will be highest right under the surface layer and then will decrease as the stress decreases with increasing distance from the surface.

It has been stated by many investigators and can be granted as a fact that dislocations will pile up at hard inclusion particles. Dislocations represent minute centers of internal stress in a material, Honeycombe [8]; so as more and more dislocations are piled up at hard inclusion particles a stress concentration is built up at the inclusion-matrix interface. At a certain point the stress concentration from the dislocation pile-up relaxes or levels out. There are many ways in which the stress relaxation can take place. One way is that the dislocation can bypass the particle. In this case the material behaves as completely homogeneous and no voids are created. This happens where the particle size is very
small. The very high ductility of T. O. Nickel (Ni + 2.3 vol.
- % Tho), which has only ~ 0.04 μm average particle diameter
is explained by this kind of stress relaxation, Rosenfield [9].

When the particle inclusion has a size which makes it
impossible for the dislocations to pass it, the stress relaxa-
tion must happen in some other way. The way in which it hap-
pens depends on the relative strength of the inclusion-matrix
interface and the strength of the inclusion particle. If the
strength of the inclusion-matrix interface is as high or
higher than the strength of the particle, the dislocations can
move through the particle without any void formation.

On the other hand, when the strength of the particle
inclusion is higher than the inclusion-matrix interface, the
stress relaxation results in a separation of the inclusion-
matrix interface and a void will be created. In some cases
the inclusion-matrix interface can be broken before any stress
is applied. Backofen [10] indicates that the bond between the
particle and the matrix can be fractured under the stresses
set up by thermal contraction during cooling. This can only
happen where the expansion coefficient of the particle is
greater than that of the matrix and where the particle-matrix
bond is weak. Therefore a large number of alloys would not be
expected to contains such pre-existing holes.

It is well known that commercial metals generally con-
tain hard inclusions in the form of oxides, carbides, and
nitrides, etc. In view of that and what has been said above,
it can be assumed that the most critical factors in the void
creation are the dislocation density in the matrix and the inclusion-matrix interface strength. Now, as the strain deformation of the sub-surface layer decreases with an increasing distance from the surface, one could expect that the dislocation density would differ in the same fashion. This leads to the conclusion that at a certain distance from the surface, depending on the material involved, the dislocation density in the strained layer has reached the point where its buildups at inclusions are enough to break the band between particle and matrix. Or in other words, one could expect to see initiation of void at some critical distance from the surface, depending on the material involved. At a lower distance from the surface the dislocations will be absorbed by the already existing holes resulting in hole growth.

Argon [11] has introduced a model where the cavity formation around inclusions of sub-micron size is explained by continuum deformation. It is considered here that the particle is a rigid cylinder and the surrounding matrix as an elastic, plastic strain hardening continuum. The determination of the stress as the inclusion matrix interface is the basic factor. The interfacial total tensile stress is said to be:

$$\sigma_{rr} = Y(\bar{\varepsilon}^P) + \sigma_T$$

where $Y(\bar{\varepsilon}^P)$ is the flow stress in the region of the inclusion for the average local plastic strain if the inclusion had not been present. $\sigma_T$ is from an unpure shear deformation field
field which has to be added to the plastic drag. The equation shows that the interfacial stress will increase with strain hardening and with triaxiality which promotes cavity formation.

The possibility, or more correctly the necessary amount of applied deformation for void creation differs depending on the size of the inclusion and their volume concentration. At very small inclusions where no local diffuse plastic relaxation is possible as there are no dislocations in the surroundings, shear stress relaxation at the matrix-inclusion interface is explained by punching out dislocations loops. This is explained in Fig. 10. It shows inclusion with radius $\rho$. The ellipse shows the configuration which the inclusion should take under applied load if it were as soft as the matrix. But as it is assumed that the inclusion is completely rigid, tensile and compressive stresses will act at the interface. It is assumed where the tensile stress is acting that cavity is prevented by plastic punching by a cylinder with a diameter $\sqrt{2}\rho$. The tensile stress at the interface is determined by the wall friction of the cylinder. When the inclusion is small the friction force which acts on the cylinder creates a tensile stress at the interface which is less than the inclusion-matrix band and no cavity is formed. When the inclusion diameter increases ($> 500^\circ\text{A}$) the wall friction can create tensile stress which reaches the interfacial strength and a cavity is formed. When the volume concentration of the secondary particles is so small that these secondary plastic zones do not touch, the particle is isolated. On the other hand,
when volume concentration is so large that the secondary zones touch, interaction between particles occurs and the possibility of void creation increases.

3.3 Hole Growth in the Sub-Surface Layer

In the preceding chapter the creation of voids around hard inclusions was discussed. In this chapter a brief discussion on the possible mechanism of hole growth will be given.

In the last decade many electron microscope studies have been made on hole growth; e.g. Palmer [12] and Beachem [13]. Although much can be learned from these studies about hole growth, more quantitative data is needed to provide satisfactory checks for the proposed models. Rosenfield [9] shows graphically how great the difference is on the models which are used to describe hole growth. There are basically three models in the literature which describe the hole growth phenomena: the solid-mechanics model, the empirical model and the dislocation pile-up model.

McClintock's et al. [14] analysis of hole growth is based on solid mechanics. It is very reasonable as it takes into account the effects of multiaxiality of stress and work-hardening. At zero strain a hole is considered to have an ideal spherical configuration. Under an applied strain the hole deforms into an ellipsoid (See Fig. 11). The elliptical hole then rotates, grows or changes eccentricity, depending on the $\sigma/\tau$ ratio. For instance, at $\sigma/\tau = 1$ the hole elongates in the direction of principal stress. The deformation of the shear band rotates the hole and tends to close it.
The principle tensile stress works against this and tends to rotate the hole backwards and round it out. So a steady state is reached at a certain point (shear strain $\approx 5.5$). Beyond that the hole grows to constant orientation and eccentricity. When $\sigma/\tau < P$, McClintock's analysis states that the hole will close at some shear strain and then rotate as a material line as the shear strain continues to increase; e.g. for $\sigma/\tau = 0$ the hole will close at shear strain $\approx 1.2$. This is contrary to what has been observed microscopically; e.g. Gurland [15], Jahanmir [7]. In those cases the hole never closes but rotates and becomes a thin crack as the shear strain increases. While McClintock's model suggests that the hole exists at zero strain, an empirical formulation worked out by Gurland [15] suggests that holes start forming at a finite value of strain; in this case $> \sim 20\%$. This is quite in agreement with the experimental observations.

A third way of explaining the hole growth has been proposed by McLean [16]: the motion of dislocations (See Fig 9). His model is based on the fact that a region below an edge dislocation can be considered as a minute crack. Thus, when a dislocation moves around and hits an already existing hole it will be absorbed by the hole which then increases in size.

It is suggested here that the growth and deformation of holes in the strained sub-surface layer can be explained by a combination of McClintock's and McLean's models. A hole, as it gets closer and closer to the surface within the sub-surface layer, will rotate and elongate with the increasing shear.
strain. At the same time it will constantly be increasing its area by absorbing dislocations. As the shear strain right under the surface layer probably is very high, holes there will be very much elongated, with the major axis almost parallel to the surface.
CHAPTER 4

DETERMINATION OF SHEAR STRAIN IN METAL

4.1 Introduction to the Linear Intercept Method

Since metal grains never have an ideal spherical form, another method different from the one introduced in Chapter 2 is necessary to determine the effective plastic strain due to pure shear. Dautzenberg [1] has introduced a method which can be used to do this. It is called "The Linear Intercept Method" and is based on the model introduced in Chapter 2.

A linear intercept is a line which lies in the x-y plane parallel to the y-axis. Its length is the distance between the two intersections which the line makes with a grain boundary, Fig. 13. By measuring a certain number of linear intersections and using the fact that the average linear intercept possesses a constant ratio to the smallest axis of the ellipsoid arising from a sphere by true shearing, the effective strain at the grain can be determined.

4.2 Mathematical Derivation of the Linear Intercept Method

The volume $V$ of an ellipsoid is, Murray [18]

$$V = \frac{4}{3} \pi \frac{A \cdot B \cdot D}{8}$$

(15)

where $A, B, \ell, D$ are the main axis of the ellipsoid. See Fig. 14. A small areal element in the x-z plane is $dx, dz$, so the volume $V$ is also:
\[ V = \sum_{i=1}^{n} c_i \, dx_i \, dz_i \]  \hspace{1cm} (16)

or

\[ V = \Delta x \, \Delta z \sum_{i=1}^{n} c_i \]  \hspace{1cm} (17)

where \( n \) is equal to the number of small elements.

The average linear intercept \( \bar{c} \) is defined by:

\[ \frac{1}{n} \sum_{i=1}^{n} c_i = n \bar{c} \]  \hspace{1cm} (18)

The area of the ellipse with \( A \) as the major axis and \( D \) as minor axis is:

\[ \frac{\pi DA}{4} \]

The DA's plane makes the angle \( \alpha \) to the XZ plane so the projection of the small area element \( dx; dz \) on the DA's plane is:

\[ \frac{dx_i \, dz_i}{\cos \alpha} \]

and

\[ \frac{n \, dx_i \, dz_i}{\cos \alpha} = \frac{\pi DA}{4} \cdot \cos \alpha \]  \hspace{1cm} (19)

Equation (17) can be written as

\[ \frac{dx_i \, dz_i}{\cos \alpha} \cdot \cos \alpha \sum_{i=1}^{n} c_i = V \]  \hspace{1cm} (20)
Substituting Eqs. (19), (18), (15) into (20) gives

\[
\frac{D}{n^4} A \cos \alpha - \frac{4}{3} A B = \frac{D}{8}
\]

or

\[
\overline{c} \cos \alpha = \frac{2}{3} B
\]

(21)

Now a new coordinate system is introduced into Fig. 14 which has centrum at \((\frac{1}{2} D \tan \gamma, \frac{1}{2} D)\) and axis parallel to the major and minor axes of the ellipse in the xy plane. The A-axis will then go from \(J'\) to \(I'\) and the D-axis from \(N'\) to \(M'\). If this ellipse is transferred back to the original circular stage, point \(M'\) will go to \(M\), \(N'\) to \(N\), \(I'\) to \(I\) and \(J'\) to \(J\). Representing these new points in the xy system yields, \(B = D \tan \alpha\) and \(A = \frac{D}{\tan \alpha}\) .

(22)

Equations (21) and (22) give

\[
\overline{c} \cos \alpha = \frac{2}{3} D \tan \alpha
\]

and by use of Eq. (13)

\[
\overline{c} \cos \alpha = \frac{2}{3} \overline{c} \frac{\tan \alpha}{\cos \alpha}
\]

(23)

From Fig. 14

\[
\tan \alpha = \frac{1/2 D}{1/2 D \tan \gamma + DF}
\]

\[
1/2 D \tan \gamma + DF = \frac{1/2 D}{\tan \alpha}
\]
So

\[ \tan \gamma = \frac{1}{\tan \alpha} - \frac{D F}{1/2 D} = \frac{1}{\tan \alpha} - \frac{2 D F}{D} \]

\[ \tan \gamma = \frac{1}{\tan \alpha} - \tan \alpha = \frac{1 - \tan^2 \alpha}{\tan \alpha} \]

\[ \tan \gamma = \frac{2}{\tan \frac{\alpha}{2}} \quad (24) \]

and (24) into (23) gives

\[ \overline{c} = \frac{2}{3} c (1 + \tan^6 \alpha)^{1/2} . \]

For \( \alpha < \pi/8 \) that is for \( \gamma > \pi/3 \) within an accuracy of 1%

\[ \overline{c} = \frac{2}{3} c \quad . \quad (25) \]

For a sphere the following holds:

\[ V = \frac{1}{6} \pi D^3 = \pi \frac{D^2}{4} \overline{D} \]

where \( \overline{D} \) average diameter of the sphere and \( \pi \frac{D^2}{4} \) is the area of the plane which goes through the centerum of the sphere.

This gives

\[ \overline{D} = \frac{2}{3} D \]

and Eq. (14) becomes

\[ \overline{\varepsilon} = \sqrt{\left(\frac{\overline{D}}{\overline{c}}\right)^2 - 1} \quad . \]
CHAPTER 5

EXPERIMENTAL PROCEDURE

An experiment was done to measure and investigate the strain field near the surface created by surface traction. Sliding wear tests were performed on the specimen to create the surface traction. After the wear tests the specimen was sectioned, polished and etched. Pictures of the deformed layer were then taken by a scanning electron microscope. From these pictures the deformed layer was investigated and the strain field determined.

In this chapter the preparation of the specimen will be described as well as the wear test procedure and the method of obtaining data.

5.1 Specimen Preparation

As the grains were used in determining the strain field in the deformed layer, some constraints were put on the selection of materials. The reason for this is that in order to be able to get accurate data the grains have to be smaller than the thickness of the deformed layer. The grain size of the specimen chosen was about 1/5 of the thickness of the deformed layer.

A. Chemical Composition and Processing Methods

a. Doped AISI 1020 Steel

The metal used is AISI 1020 steel which is doped with tungsten, zirconium and manganese to give it a fine grain structure. It has a carbon content of 0.17%. Other alloying
elements were 0.6\% Mn, 0.05\% W and 0.25\% Zr. The tungsten and zirconium additions retards grain growth and raise the recrystallization temperature of the iron.

The steel was melted by arc melting in an inert atmosphere. It was hot rolled, followed by cold rolling to one inch diameter. The one inch rod was then light machined down to 5/8 inch diameter to remove hot or cold working marks.

b. Copper

The copper had some alloying elements to inhibit grain growth which also gave a better grain structure when etched. The composition of the copper was: 99.5\% Cu, 0.25\% Cr, 0.10\% Ti and 0.08\% Cb.

The copper was received as a 1/4 x 1/2 inch cold rolled bar. Then a rod .235 inches in diameter was made from the bar by machining.

B. Preparation for Wear Tests

To impose surface traction on the specimen an unidirectional wear test was conducted. The preparation of the specimen was done in the following way.

a. Specimen Preparation for Wear Tests

A lathe was used to get the unidirectional motion so that all specimens had cylindrical shape. The AISI 1020 specimen had a 5/8\" diameter when it was received and was only given a surface finish by a grade 1 emery paper. The copper specimen was machined from 1/4\" x 1/2\" bar to 1/4\" circular bar. Its surface finish was the same as that of the AISI 1020
specimen. The length of both specimens was approximately 2-1/2 inches. After surface treatment all specimens were washed with soap and hot water, rinsed with alcohol and dried by a jet of air.

After cleaning, the copper specimen was annealed in a high vacuum furnace (2 x 10^{-6} - 2 x 10^{-5} \text{ mm Hg}) to minimize oxidation. Annealing time and temperature were chosen to produce minimum grain size after recrystallization (See Table I for annealing procedures). Before wear test, both specimens were immersed in trichloroethylene for 10 minutes to insure removal of all possible grease from the surface.

b. Preparation of the Slider

The same kind of slider was used in both wear tests. It was made of AISI 52100 steel with 560 kg./mm² Brinnel hardness. It was cylindrical with a 1/4 inch diameter and its length was 2-1/2 inches. The cylindrical surface was used for the sliding test. When a new test was conducted the slider was just rotated so that a new area was in contact with the specimen. Before each test the slider was immersed in trichloroethylene for degreasing.

5.2 Testing Equipment

A cylinder-on-cylinder sliding geometry (Fig. 15) was performed on a lathe. As shown in Fig. 16 the specimen was held by the rotating spindle and the slider was stationary. The cylindrical slider was mounted on the slider holder so that it could be rotated to a new position by unscrewing two
stop screws. See Fig. 17. The slider holder was connected to a lathe tool dynamometer by a steel bar. The dynamometer was attached to the carriage of the lathe. The normal load applied to the cylinder was done by the transverse motion of the carriage and recorded on a Sanborn 50 Recorder through the dynamometer. The dynamometer also picked up the friction force which was recorded on the Sanborn recorder.

During the wear test the sliding interface was flooded with argon gas at a flow rate of ten liters per minute. The spindle rotation speed and the applied normal load for each test is shown in Table. All tests were done at room temperature.

5.3 Specimen Preparation for Microscopic Observations

A. Sectioning of Specimen

In order to observe the deformed layer under the wear track the specimen was sectioned through approximately the middle of the weartrack for metallographic observations. Before sectioning the distance from the "free" end to the "outer" edge of wear track as well as to the middle of the wear track was measured using a toolmaker micrometer (See Fig. 18). This was done so that the location of the cross-section area could be known with respect to the middle of the wear track under the grinding and polishing process. An attempt was made to have the final cross-section area within 0.01 inch distance from the wear track's centrum.

The worn specimen was cut by a hack saw approximately 1/8 inch away from the wear track, parallel to the sliding
direction and perpendicular to the worn surface (see Fig. 19). Then the specimen was placed in a grinding machine and ground carefully down to the edge of the wear track. The next step was to place the specimen in a special holder (see Fig. 20) and abrade it on grade 1 paper. When the desired location was reached the final abrasion was performed on 320 and 600 grid silicon carbide papers, respectively. A metallographic polishing wheel was used for the final polishing. It was impregnated with alumina dust and the size of the alumina particles was 0.3 μm and 0.05 μm, with 0.05 μm as the final polishing particle size.

B. Protection of the Edges

As it was very important to be able to see the dimensions of the surface grains, protection of the edges became the most critical factor in the experiment.

Clear cast and Bakelite molded around the specimen gave unsatisfactory results. As the molding materials were much softer than the specimen, a rounding of edges was observed, preventing observation of the surface grains. Plating of the worn surface was not tried because of the danger that the necessary acid cleaning would damage the surface layer. Since former attempts to use copper plating followed by chromium electroplating in similar circumstances [7] had not given a satisfactory result, it was not tried here. The best result was achieved by using specimen holder as is shown in Fig. 20. It was made of aluminum. A 1/2 inch piece was cut from a circular bar, one inch in diameter. In the middle of the
holder a hole was made for the specimen. To get the best protection of the edges the hole had to be as small as possible or so small that a light force was needed to push the specimen into the hole. This was necessary as every additional difference in hole and specimen size results in an increasing gap between them. Too much gap means a rounded effect at the edges.

C. Etching and Coating

After the final polishing operation the specimens were removed from the holders, washed with tap water, dried and etched to reveal the microstructure. A 2% Nital solution was used to etch the AISI 1020 specimen. The copper specimen was polished and etched electrolytically in a solution of Orthophosphoric acid, Smithells [19].

After etching, the microstructure was observed in an optical microscope to check the effectiveness of the etching procedure. In order to insure as good results as possible when the deformed layer was observed by scanning electron microscope at the highest magnifications (20,000X), the surface was coated with 50-100 Å thick layer of gold. The gold coating was done by vapor deposition.

Table I gives a summary over the materials used, their composition, heat treatment, metallurgical history, hardness, grain size and etching reagents.
5.4 Measurement Equipment and Method

A. Preparation of Film for Measurements

Pictures of the specimen were taken on a Polaroid film type 52 or 55. The pictures were taken showing the surface edge, the deformed layer and some portion of the undeformed matrix. To do this, and at the same time get a reasonable magnification of the grains, a series of pictures had to be taken. Afterwards the pictures were cut at the appropriate places and assembled. After assembling, the pictures showed a continuous sequence from the surface into the undeformed region.

When the pictures had been assembled, coordinate axes were introduced to them with the y-axis along the surface edge and the x-axis perpendicular to it. Marks were put on the x-axis, each representing some distance from the surface in \( \mu m \). To be able to observe the grains closest to the surface, a picture of higher magnification was taken of that same area. Y-and-x axes were marked on that picture in the same way.

After this, the picture was placed under a camera and duplicated on a slide film, Polaroid type 46/146. Then the slide was placed between two glass plates.

B. Measurements

For measurements the glass plates with the slides were placed in an Optical Comparator (see Fig. 3). The grain structure was very clearly projected from the slide on the comparator's screen. The slides were adjusted in such a way that the x-axis of the picture was parallel to the comparator's table and its movable directions.
The measurement of the linear intersections was done in the following way. First, the comparator's table was moved so that the intersection of the screen axis fell on the x-axis at the distance from the y-axis where the average linear intersection was to be calculated. Then the table was lowered a certain distance so the screen axis' intersection moved parallel to the y-axis. The linear intersection at that point was then measured by moving the table in the x-direction so the intersection of the screen's axis fell on the "upper" grain boundary. Then the table was moved until it fell on the "lower" grain boundary. The x-coordinates of each point was read from the micrometer which moves the table. The difference between the x-coordinates then gave the linear intersection. Then the table was lowered the same distance as before in the y-direction and the linear intersection measured as before, etc. (see Fig. 21). The distance between intersections did depend on the average size of the undeformed grains. It was estimated that at the average, four linear intersections were taken at each undeformed grain. At least 25 linear intersections were measured at each distance from the surface. To get the average linear intersections for the undeformed grain, at least 75 measurements were used.
CHAPTER 6

EXPERIMENTAL RESULT

The experimental result will be presented in this chapter. It consists of pictures which were taken of the specimen and determination of the strain field in the deformed layer. Void and crack formation were also studied. The selection of the AISI 1020 steel and the lightly alloyed copper specimen was done to compare the relative difference in hardness and the corresponding strain field between those two materials. Table 2 gives summary of experimental conditions. It consists of specimen dimensions, normal load, friction coefficient and sliding velocity.

6.1 Determination of Strain Field

When the deformed layer had been observed and a picture taken of it, the strain field was calculated as described in sections 4.2 and 5.4.

A. Doped AISI 1020 Steel

Figures 22 and 23 show pictures of the deformed layer taken on a scanning electron microscope. They show the surface edge and extend down to the undeformed matrix. It can be seen very clearly from those pictures how the grain thickness decreases as they get closer to the surface. The better quality of the picture in Fig. 22 is due to the fact that the specimen was coated with gold before observation.

As explained in sections 4.1 and 4.2 the grain thickness indicates the amount of shear deformation. The pictures
therefore indicate that the depth of deformation is ca. 45\( \mu \)m. Also that the strain rate is highest closest to the surface.

Results from the measurement of the effective strain at different distances are represented in Fig. 25 where the effective strain is plotted versus distance from the surface. This graph shows that at a distance 1\( \mu \)m from the surface the strain is as high as 14.3 and that from the surface down to ca. 15\( \mu \)m there is a region of very high strain rate. It was technically impossible to determine the strain at the surface as the grain boundaries became invisible at high magnification. In order to get some idea about what the shear strain could be expected to be at the very surface, the three data points closest to the surface were assumed to lie on a straight line. This approximation indicated that the effective strain at the surface was ca. 16.5. The other portion of the curve was assumed to be an inverse function of \( x \). Figure 26 shows the data points for \( x > 4.14 \) plotted in log. vr. log scale. The equation for the effective strain for \( x < 4.17 \) turned out to be

\[
\bar{\varepsilon}(x) = 16.5 - 2.2x
\]

and for \( x > 4.17 \),

\[
\bar{\varepsilon}(x) = 44 \cdot x^{-1.26}
\]

A graph of these equations is shown in Fig. 25.
B. Copper

In Figs. 27(a) and (b) pictures of the deformed layer and the undeformed matrix in copper are shown. They show that the deformed layer is about 90μm deep. The experimental data from those pictures is plotted in Fig. 28. It can be assumed from the curve shown there that the effective strain at 10μm depth is ca. 12. It turned out to be extremely difficult to obtain data for distance smaller than 10μm from the surface. The reason for this was that the size of the grains closest to the surface became very unclear. This can be seen in Fig. 24 which is a higher magnification (2000x) of the surface layer in Fig. 27. Reliable data is therefore not at hand so that the effective strain at the very surface could not be determined. Another difficulty involving the strain calculations in the copper specimen is the great variation in grain sizes. This means that extensively more measurements have to be taken for each data point than in the case of the AISI 1020 steel specimen.

6.2 Void Formation

The deformed sub-surface layer was studied to get some idea about at what strain and under what conditions voids are formed. It was observed that a void only forms at a hard particle. In some pictures a void without any inclusion can be seen. In these cases the inclusion most probably has been there but been removed under the polishing process.
A. Doped AISI 1020 Steel

Voids seem to start at inclusions with small interfacial spacing and location at grain boundary. This can happen at as much as ca. 25μm distance from the surface and at ca. 0.8 effective strain (see circle C in Fig. 22). As a matter of fact, no voids were observed at inclusions located inside a grain. The reason for this may be that the interfacial strength between the grain material and the inclusion is stronger than between the material in the grain boundary and the inclusion. In all the pictures the effect of inclusion interaction was obvious. See circle B and C in Fig. 22 and Fig. 29(A) and (B).

In Fig. 29(b) (circle A and B) and Fig. 29(a), which gives magnification of circle B, still another form of void creation and hole growth can be observed. This is when inclusions are so close together that they touch each other. When the matrix is sheared the inclusions separate, leaving a gap between them. The effectiveness of this kind of hole formation and growth is schematically illustrated in Fig. 30. In Fig. 30(a) the inclusions are shown at zero strain and assumed to be parallel to an axis which is perpendicular to the surface. Fig. 30(b) shows the relative placement of the inclusion when the matrix has been sheared the angle \( \gamma = 64^\circ \) which is the angle of shear at 20μm depth. The effective strain at 20μm is ca. 1.2 and Eq. (8) gives \( \gamma \approx 64^\circ \). The particle spacing or the length of the created void is determined by the equation.
\[ L = D \left( \frac{1}{\cos \gamma} - 1 \right) \]

(\(\gamma\) is determined by the Eq. \(\tan \gamma = \sqrt{3}\), see chapter 2.2,)

Eq. [8]) where \(L\) is the void length, \(D\) particle diameter and \(\gamma\) angle of shear. This shows how effective this kind of hole formation and growth is; e.g. at 10\(\mu\)m from the surface, the effective strain is ca. 3.5 which gives \(\gamma \approx 80^\circ\) and \(L\) becomes \(D \cdot 4.76\).

The observations revealed that the void density increased with increased deformation (see Fig. 31).

**B. Copper**

When the copper specimen was investigated quite a different story came up. Rarely, a void could be seen. The reason is the low inclusion density in the copper. Figures 27(a) and (b) show how small the inclusion density is and also how uniform the inclusion distribution is; e.g. the average particle spacing was 23\(\mu\)m which is about 15 times the average particle diameter. All this decreases the possibility of void formation.

### 6.3 Crack Formation

**A. Doped AISI 1020 Steel**

Crack formation in the steel specimen always seemed to be associated with an existence of a hole. The creation of the crack happens basically in two different ways: gradually or instantaneously. The crack in Fig. 32 is believed to have
formed gradually. A group of inclusions staying close together has formed holes by the process described in Fig. 30 as the matrix around them has been deformed.

The other mechanism of crack formation is demonstrated in Fig. 33. It shows two voids which have formed around two neighbor inclusions (one of the inclusions has apparently been removed under the polishing operation). By looking carefully one can observe that a crack has formed between the two voids. This crack has most likely formed instantaneously by fracture of the highly sheared material (ε \approx 9.3). The cracks shown in Fig. 11 and 12 will then elongate as the distance from them to the surface decreases.

B. Copper

Figure 34(b) shows different stages of crack formation in copper. It shows (arrow A) a very narrow and long crack which lies approximately parallel to the surface. It also shows another crack (arrow B) which is closer to the surface than the first one. This crack has widened and is well visible. Figure 34(a) also shows that at one place one end of a crack has extended to the surface and a wear sheet is to be formed (arrow C). Figure 34(b) shows a place where a crack has extended to the surface and a wear sheet is forming. At the same time it shows that the wear sheet itself is filled up with cracks which indicates that the sheet will break up into smaller particles. Finally, Fig. 34(c) shows a wear sheet formation.
The mechanism of the crack formation in copper is not as obvious as in the case of the AISI 1020 specimen where the cracks always were associated with voids formed around inclusions. In copper where the inclusion density is so small it is not as easy to relate the crack formation to the existence of holes. It is believed though that the few holes which are formed are the initiation of the cracks. The crack elongation is then an accumulation of incremental fracture which takes place each time the slider passes. The crack in the copper specimen seemed to start at a depth as much as 40–50 µm and at effective strain ca. 2.0.
CHAPTER 7
DISCUSSION

7.1 Microscopic Observations

The microscopic observations clearly reveal how important a part the second phase particles play in the mechanism of sliding wear. The observations also show how strain field created by surface traction differs depending on the hardness of the material. The many possible mechanisms of creation of the observed voids and cracks have already been discussed and will not be repeated here. It should be noted that the material investigated here is specially treated to contain controlled amount of inclusions. Therefore, one could expect to see many more inclusions in the commercial materials. Or furthermore, the material could contain pre-existing holes resulting from processing operation. The processing operation may also have caused unfavorable placement of the inclusions; e.g. stored them parallel to the surface and therefore increased the possibility of crack formation.
The experimental result shows that strain due to pure shear can be determined very accurately by observing the distortion of an idealized spherical grain.

This can be applied to actual grains by using the average linear intersection of the deformed and undeformed grains. The experimental result shows that in doped AISI 1020 steel the thickness of the deformed layer is as much as 45μm and the effective strain at the surface is 16.5. The region of high strain gradient is from the surface down to 1/4 of the deformed layer.

In an alloyed copper (0.25% Cr, 0.10% Ti, 0.08 Cb) the thickness of the deformed layer becomes much greater or 90μm, and apparently the effective strain at the surface is 5 to 6 times that in the case of AISI 1020 steel.

The microscopic observations show how nucleated voids are around hard inclusions due to the high deformation. In the case of the steel this happens at effective strain. As the deformation increases the voids become bigger and elongate. Then the voids link together and form cracks. The cracks again elongate to form wear sheets.
REFERENCES


References (continued)


LIST OF TABLES

1. Specimen description.
2. Experimental conditions.
LIST OF FIGURES

1. Deformation of an ideal spherical grain under pure shear.
2. Test device to impose a pure shear deformation on plastine specimen.
3. Optical comparator.
4. Grid system to locate the centrum of the ellipse.
5. Plastine specimen before deformation.
6. Plastine specimen after deformation.
7. Specimen dimensions.
8. Sketch to show the relative position of: movable bar, circle on specimen and the reference axis.
9. Schematic representation of the dislocation model.
   (a) dislocation and edge of a hole.
   (b) dislocation approaching a hole.
   (c) dislocation absorbed by a hole resulting in hole growth.
10. Argon's model of void formation.
11. McClintock's model of hole growth.
12. Holes right under the surface layer.
13. Linear intercept.
14. Sphere deformed under pure shear.
17. Slider holder.
18. Measurements by toolmaker micrometer before sectioning of the specimen.
20. Specimen holder.

51
List of Figures (continued)

21. Method of measuring the linear intersections at one distance from the surface.

22.- Deformation in AISI 1020 steel.

23. Deformation in copper close to the surface.

24. Strain field near the surface due to surface traction in doped AISI 1020 steel.

25. Data from the AISI 1020 specimen plotted in log vr. log scale.

26. Deformation in copper (99.57% Cu, 0.25% Cr., 0.10% Ti., 0.08% Cb.).

27. Strain field near the surface due to surface traction. Copper (99.75% Cu., 0.25% Cr., 0.10% Ti., 0.08% Cb.).

28. Hole formation around inclusion in steel.

29. Schematic representation of void formation and hole growth when matrix with two touching inclusions is sheared.

30. Void density increases with decreasing distance from the surface.


32. Crack formed by fracture.

33. Crack and wear sheet formation in copper.
<table>
<thead>
<tr>
<th>Metal</th>
<th>Chemi composition %</th>
<th>Heat treatment and processing</th>
<th>Brinnel hardness kg/mm²</th>
<th>Etching</th>
</tr>
</thead>
<tbody>
<tr>
<td>Doped AISI 1020 steel, specially prepared.</td>
<td>.17 Cr.</td>
<td>Hot and cold rolled, annealed at 680°C for 17 hours and 600°C for 14 hours.</td>
<td>159</td>
<td>2% Nital</td>
</tr>
<tr>
<td></td>
<td>.60 Mn.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>.50 W.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>.25 Zr.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Copper</td>
<td>99.57 Cu.</td>
<td>1/4 x 1/2 hot rolled bar, annealed at 250°C for 3 hours.</td>
<td>76</td>
<td>Electrical solution:</td>
</tr>
<tr>
<td></td>
<td>.25 Cr.</td>
<td></td>
<td></td>
<td>Orthophosphoric acid.............200cc</td>
</tr>
<tr>
<td></td>
<td>.10 Ti.</td>
<td></td>
<td></td>
<td>H₂O...............100cc</td>
</tr>
<tr>
<td></td>
<td>.08 Cb.</td>
<td></td>
<td></td>
<td>5 amp/dm² 1.8-2.0 volts, temperat. room</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Electrode horizontal, Specimen lower electr. No stirring.</td>
</tr>
<tr>
<td>Specimen</td>
<td>Normal load, kg</td>
<td>Frictional force, kg</td>
<td>Friction coefficient</td>
<td>Diameter cm</td>
</tr>
<tr>
<td>---------------</td>
<td>-----------------</td>
<td>----------------------</td>
<td>----------------------</td>
<td>-------------</td>
</tr>
<tr>
<td>Doped AISI 1020 steel</td>
<td>2.4</td>
<td>1.8</td>
<td>0.76</td>
<td>1.6</td>
</tr>
<tr>
<td>Copper</td>
<td>2.1</td>
<td>1.3</td>
<td>0.64</td>
<td>0.60</td>
</tr>
</tbody>
</table>
Fig. 1. Deformation of an ideal spherical grain under pure shear.
Fig. 2. Test device to impose a pure shear deformation on a plasticine specimen.
Fig. 3. Optical comparator

Fig. 4. Grid system to locate the centrum of the ellipse.
Fig. 5. Plasticine specimen before deformation

Fig. 6. Plasticine specimen after deformation
Fig. 7. Specimen dimensions
Fig. 8. Sketch to show the relative position of: Movable bar, circle on specimen and the reference axis.
Fig. 9. Schematic representation of the dislocation model.

(a) dislocation and edge of a hole

(b) dislocation approaching a hole

(c) dislocation absorbed by a hole resulting in hole growth
Figure 10. Argon's model of void formation.
Fig. 11. McClintoc's model of hole growth.

Fig. 12. Holes right under the surface layer.
Fig. 13. Linear intercept.

Fig. 14. Sphere deformed under pure shear.
Fig. 15. Unidirectional wear test. Cylinder-on-cylinder geometry.

Fig. 16. Testing equipments.
Fig. 17. Slider holder.

Fig. 18. Measured before sectioning by microscope micrometer.

a. The distance from the "free" end to the middle of the wear track.

b. The distance from the "free" end to the edge of the wear track.
Fig. 19. Schematic representation of specimen preparation for observation.

Fig. 20. Specimen holder.
Fig. 21. Method of measuring the linear intersections at one distance from the surface.
Fig. 22. Deformation in AISI 1020 steel.
Fig. 23. Deformation in AISI 1020 steel.

Fig. 24. Deformation in copper close to the surface.
Fig. 25. Strainfield near the surface due to surface traction. Doped AISI 1020 steel.
Fig. 26. Data from the AISI 1020 specimen plotted in log. vs. log. scale.
Fig. 27. Deformation in copper (99.57% Cu, .25% Cr, .10% Ti, .08% Cb).
Fig. 28. Strainfield near the surface due to surface traction. Copper (99.57% Cu., .25% Cr., .10% Ti., .08% Sn.).
Fig. 20. Hole formation around inclusion in steel.
Fig. 30. Schematic representation of void formation and hole growth when matrix with two touching inclusions is sheared.
Fig. 31. Void density increases with decreasing distance from the surface.

Fig. 32. Gradual crack growth.
Fig. 33. Crack formed by fracture.
Fig. 34. Crack and wear-sheet formation in copper.