

STUDY OF THE VARIATION OF MECHANICAL PROPERTIES
WITH RESPECT TO GRAIN SIZE IN HYPOEUTECTOID STEEL

By

A.A.MD.REZAUL HAQUE

A thesis submitted to the Department of Metallurgical Engineering, Bangladesh University of Engineering and Technology, Dacca, Bangladesh in partial fulfilment of the requirements for the degree of Master of Science in Engineering (Metallurgical).

Bangladesh University of Engineering & Technology, Dacca.

August, 1980.



(1)

CERTIFICATE

This is to certify that this work has been done by the author under the supervision of Dr. Ehsanul Haque, Assistant Professor, Department of Metallurgical Engineering, Bangladesh University of Engineering and Technology, Dacca and it has not been submitted elsewhere for the award of any other degree or diploma.



E. Haque 21.8.80.
Signature of the
Supervisor

Ehsanul Haque 21.8.80
Signature of the author.

Accepted as satisfactory for partial fulfilment of the requirements for the degree of Master of Science in Engineering (Metallurgical):

1. E. Haque 21.8.80
Dr. E. Haque Chairman
Assistant Professor,
Department of Metallurgical Engineering
BUET., Dacca.
2. M. Ibrahim
Dr. M. Ibrahim Member
Professor & Head,
Department of Metallurgical Engineering
BUET., Dacca.
3. M. Ashraf Hussain Khan
Dr. M. H. Khan Member
Professor & Head,
Department of Mechanical Engineering,
BUET., Dacca.
4. S. Islam
Dr. S. Islam Member
Associate Professor,
Department of Metallurgical Engineering,
BUET., Dacca.
5. E. Akanda
Dr. E. Akanda Member
Ex-Senior Executive,
BIDC, Dacca.



ACKNOWLEDGEMENT

The author is deeply indebted and much obliged to Dr. Elsanul Haque, Assistant Professor, Department of Metallurgical Engineering, B.U.E.T., Dacca for his guidance and encouragement in carrying out the research work as well as in writing this thesis. The author is grateful to Dr. M. Ibrahim, Professor and Head, Department of Metallurgical Engineering for his kind suggestions.

The author express his thanks to all the staff of B.U.E.T., Dacca who helped in different stages of this project.

BUET., Dacca
August, 1980.

The author..*

SYNOPSIS

Grain size is one of the main controlling factors of mechanical properties (e.g. strength & ductility) of metals and alloys at room temperature. In Bangladesh the bulk of steel used and produced is mild steel containing carbon 0.2%-0.35%. But the data on the effect of grain size on mechanical properties of this type of steel is not available. For this reason mild steel containing 0.35% C was chosen for this work. Different grain sizes were obtained by annealing the specimens at different temperatures. The micro photographs were taken and they were standardised by matching with ASTM (American Society for Testing Materials) standard grain sizes. Then their hardness and yield strength were determined. True tensile strength (instantaneous) is difficult to measure. So instead of ultimate tensile stress, yield stress was measured which is also vitally important for engineering purposes and which can be determined easily for mild steel. Relationships of grain sizes with tensile and fracture properties were obtained.

CONTENTS

	<u>Page</u>
Chapter 1 INTRODUCTION	1
Chapter 2 GRAINS AND THEIR FORMATION	4
2.1 Grains	4
2.1.1. Crystal Structure	4
2.1.2. Single Crystal & Polycrystalline Metals	4
2.2. Grain Growth	7
2.2.1. Grain boundary mobility	7
2.2.2. Law governing grain growth	13
Chapter 3 PLASTIC DEFORMATION OF METALS	20
3.1. Effect of Grain Size on Yield Point of Metals	20
3.2. Fracture Mechanism & Grain Size	23
Chapter 4 EXPERIMENTAL PROCEDURE AND RESULTS	27
4.1. Production of Austenitic Grain Size	27
4.2. Measurement of Grain Size	28
4.3. Material used and Results Obtained	31
Chapter 5 DISCUSSION AND CONCLUSION	47
Appendix 1	59
Appendix 2	62
References	63

Chapter I

INTRODUCTION



Almost all books of material science narrate that grain sizes and grain size numbers have direct relationship to the mechanical properties of metals and alloys. So it is extremely useful to have such data for a mild steel containing carbon 0.3 - 0.35%, which the Chittagong Steel Mills Ltd. (C.S.M.) is now producing. At present C.S.M. is very much interested to have this relationship. They are having some problems in fabricating ingots which may be due to the variation in grain sizes. The Bengal Rolling Mills and other such factories also faced similar problems using raw materials from C.S.M. (See appendix -2). In this work an attempt has been made to relate some of these mechanical properties with grain sizes so as to give a better understanding for their fabrications and uses. For this purpose mild steel containing 0.35% carbon which is a product of Chittagong Steel Mills Ltd. has been chosen.

Similar studies were carried out by a number of workers. The most fundamental work was done by Hall^{(1)*} and Petch⁽²⁾ and they established that the yield stress (σ_y) in iron and steel is related to the grain size (d) by the relationship.

$$\sigma_y = \sigma_0 + kd^{-1/2}$$

This relationship has been extensively studied by Petch and his co-workers⁽³⁻⁸⁾ and has proved to be a powerful equation for investigating the mechanical properties of mild steel and other b.c.c. metals at the yield point.

* Figure within parenthesis indicates the reference number.

Petch⁽²⁾ explains this equation by the formation of arrays of dislocations piled up against grain boundaries. The first term, σ_0 , appears as the stress needed to drive a free dislocation against the resistance of the Peierls-Nabarro lattice friction stress and against the resistance of dispersed impurities, precipitates and lattice defects; k is taken to be a measure of the shear stress required to unlock a dislocation.

Other explanations of the Petch equation have been proposed. Johnson⁽⁹⁾ suggested that it is more difficult to propagate a Luders band through a polycrystal than through a single crystal because the regions near the grain boundaries in a polycrystal have to be strongly work-hardened as the Luders band proceeds. Crussard⁽¹⁰⁾ suggests that free dislocations are emitted from grain-boundaries and this leads to Petch equation taking into account that the dislocation stress field is proportional to $d^{-1/2}$. Li⁽¹¹⁾ explains the Petch relation from a consideration of grain boundary sources of dislocations, and examines the effect of the angle of misfit of grain boundaries, of impurities, of temperature and plastic strain on the Petch slope.

Petch and his co-workers⁽³⁻⁸⁾ performed experiments at room temperature, determined the values of σ_0 and k and then confirmed Hall-Petch equation for plain carbon steels contain carbon 0.02%, 0.03%, 0.04%, 0.05%, 0.115%, 0.12%, 0.13% and 0.15%.

Armstrong et al.⁽¹²⁾, Jones and Holland⁽¹³⁾ used plain carbon steel containing 0.12%C and 0.2%C respectively and their result

closely agreed with Hall-Petch equation. Tjerkstra⁽¹⁴⁾ and Jaoul⁽¹⁵⁾ studied decarburised iron and also verified Hall-Petch equation.

In the following sections the grain formation and its growth have been discussed and this is followed by a discussion of the relationship between the grain sizes and the mechanical properties. The experimental technique used has been explained, finally experimental observations are presented and discussed.

Chapter-2

GRAINS AND THEIR FORMATION

2.1 Grains:

2.1.1. Crystal Structure

Metals are made up of regular arrays of atoms on simple lattices.⁽¹⁶⁾ It will be convenient to regard the individual atoms as tiny spheres with diameters of the order of 10^{-8} cm. The atoms in most metals are clustered together as closely as possible. Atomic patterns or crystal structures may be represented by three dimensional figures, such as cubes and hexagonal prisms etc. .

The majority of pure metals have one of the three relatively simple crystal structures⁽¹⁷⁾ and are known as the face-centred cubic (fcc), the hexagonal close-packed (hcp) and the body-centred cubic (bcc) structures (Fig.2.11). These figures indicate the positions of the centres of the atoms and show conveniently the unit of the pattern. It must be understood that the atoms vibrate about these positions; the amplitude increases with the rise in temperature; at room temperature it is of the order of a tenth of the atomic diameter. .

2.1.2 Single Crystal and Polycrystalline Metal.

In the simplest form, the atomic pattern is continuous throughout the whole mass of the metal. When this occurs the metal is said to be a single crystal; such a state is uncommon. Usually the pattern is discontinuous and changes its orientation from place to place (Fig. 22). Thus pure metal normally consists of a number of zones of diffe-

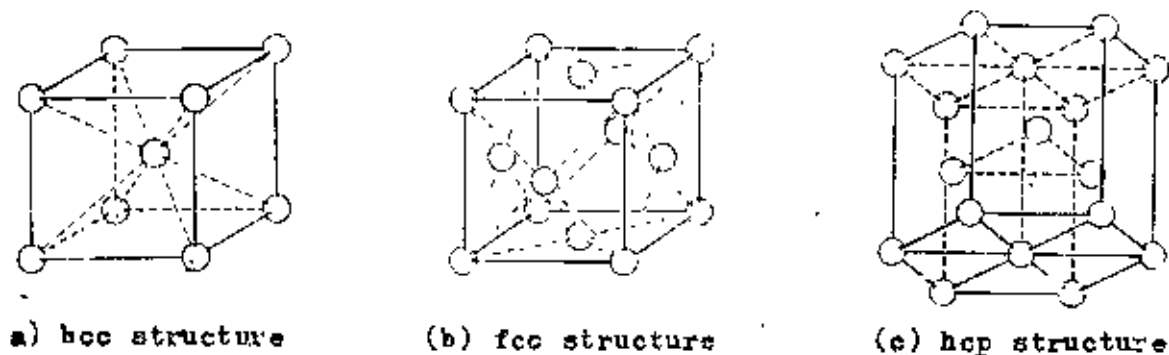
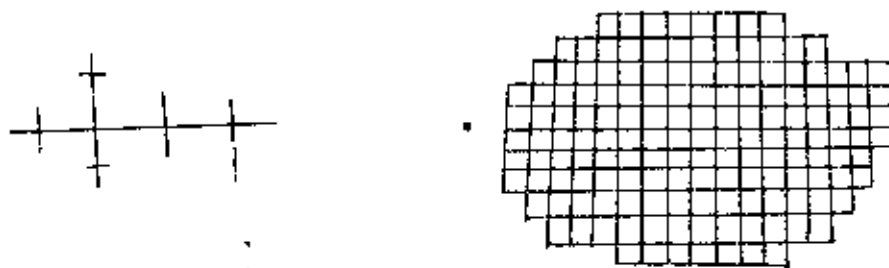
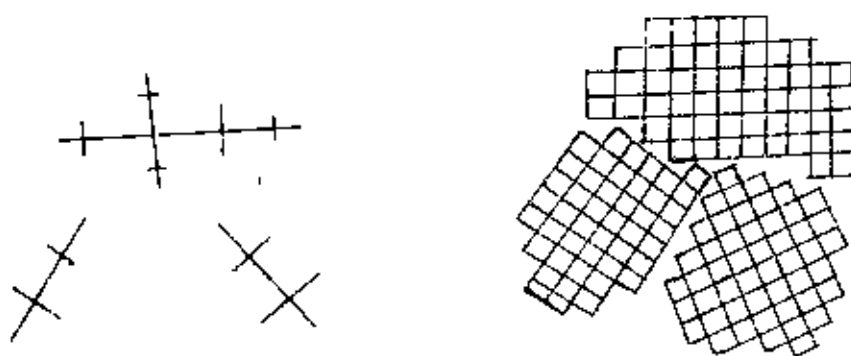


Fig.2.1 Crystal structure of pure metals.



(a)



(b)

Fig.2.2. Sketch illustrating difference between the atomic arrangement in a) Single crystal and b) Polycrystalline metal;

rent orientations, although the actual patterns in each zone is the same. The orientation may be random, or there may be preferred orientation. The zones are called crystals or grains and the metal is known as polycrystalline. The change from one orientation to another occurs over relatively small distances, and it is apparent that in these transition regions the pattern is distorted Fig. 2.2(b). The regions are known as the grain or crystal boundaries and have a marked effect on the properties of a metal.

The mismatch of the orientation of adjacent grains produces a less efficient packing of the atoms along the boundary⁽¹⁷⁾. Thus the atoms along the boundary have a higher energy than those within the grains. This accounts for the more rapid etching along the boundaries. The higher energy of the boundary atoms is also important for the nucleation of polymorphic phase changes. The lower atomic packing along the boundary favours atomic diffusion and mismatch between adjacent grains interferes with the progression of dislocation movements.

The concentration of impurities at grain boundaries makes it difficult to determine their true melting point. However it is to be expected that these boundaries will have a slightly lower melting point than that of the grain bodies. Metal in the single-crystal state (in the absence of grain boundaries) is usually softer and weaker than when polycrystalline; hence generally its only use is for experimental purposes.

The polycrystalline condition results because crystallization usually starts from a number of centres in the bulk of the metal. (Fig. 2.2(b)). Thus a different crystalline region forms from each nucleus⁽¹⁶⁾. This is true not only of crystallization from the liquid state, but also when a new set of crystals is formed in the solid state by the process of recrystallisation. The grains or crystals formed in these ways are said to be allotriomorphic, that is without any regular shape. The form of each grain is determined as a balance between its growth and that of its neighbours. Geometrically, the grains may be described as polyhedra.

2.2 Grain Growth:

2.2.1. Grain Boundary Mobility:

It is now generally recognized that the average grain size of a metal will increase with time if the temperature is such as to give significant atomic movements. The driving force for grain growth lies in the surface energy of the grain boundaries⁽¹⁸⁾. As the grain grows in size and their number decreases, the grain boundary area diminishes and the total surface energy is lowered accordingly. The growth of cells in a foam of soap also occurs as a result of a decrease in surface energy; the surface energy of the soap film. Because a number of complicating factors that influence the growth of the metal crystals do not apply in the case of soap film, the growth of soap bubbles may be taken as a rather ideal case of cellular growth. For this reason the growth of soap cells will be considered before the more complicated case of metallic grain growth is taken up.

In a single spherical soap bubble the gas enclosed by the soap film is always at a greater pressure than that on the outside of the bubble soap film. This pressure difference between the inside and outside of the soap bubble can be expressed in the form of a simple equation (18)

$$\Delta p = \frac{4Y}{R} = \frac{8Y}{D} \quad (2.1)$$

where Y is the surface tension of one surface of the film (soap films have two surfaces), R is the radius of the soap bubble and D is its diameter. This equation shows clearly that the smaller the bubble, the greater the excess pressure inside the bubble.

Because of the pressure difference that exists across a curved soap film, gaseous diffusion occurs, the net flow occurring through the film from the high- to the low pressure side. In other words, the atoms diffuse from the inside to the outside of the bubble, resulting in a decrease in size of the bubble and a movement of the walls inward i.e. toward its centre of curvature.

The above discussion can now be carried over to the more general example, a soap froth. In the froth, the cells contain curved walls, and the curvature varies from cell to cell and within the film surrounding any one cell, depending on the relative size and shape of the neighbouring cells. In all cases, however, a pressure difference exists across each curved wall, with the greater pressure on the concave side. Gas diffusion resulting from this pressure difference, in turn, causes the walls to move but always in a direction

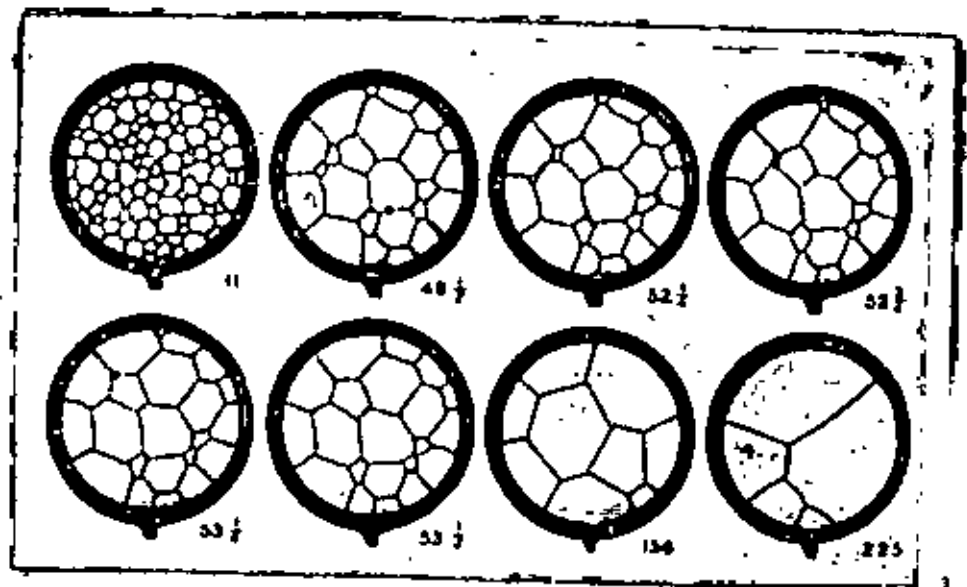


Fig. 2.3 Growth of soap cells in a flat container.



toward their center of gravity i.e. curvature!

It will be useful to find out how the movement of cell walls causes the cells in the soap froth to grow in size. For the sake of simplicity, a network of two dimensional cells, those with walls perpendicular to the plane of view may be considered. A froth of this type can be formed between two closely spaced parallel plates of glass. This simplification greatly reduces geometrical complexity, and still permits the more important principles of cellular growth to be observed. Figure 2.3 presents a sequence of pictures from the work of C.S. Smith⁽¹⁹⁾ showing the growth of the cells in a two dimensional soap froth formed in a small flat glass cell. The figure at the lower right hand corner of each photograph represents the number of minutes from the time that agitation of the cell to form the froth was ended. It also represents the time during which cell growth has taken place. In several of the photographs, small, three-sided cells can be observed. An enlarged sketch of one of these cells is shown in Fig. 2.4. It can be noted that in order to maintain the equilibrium angle of 120° required when three surfaces with equal surface tensions meet at a corner junction, the cell walls of the three-sided cell have been forced to assume a rather pronounced curvature. Since this curvature is concave toward the centre of the cell, the cell can be expected to migrate, thus decreasing the volume of the cell and causing it to disappear completely.

Further study of the photographs of Fig. 2.3 reveals that cells with less than six sides have walls that are primarily concave toward



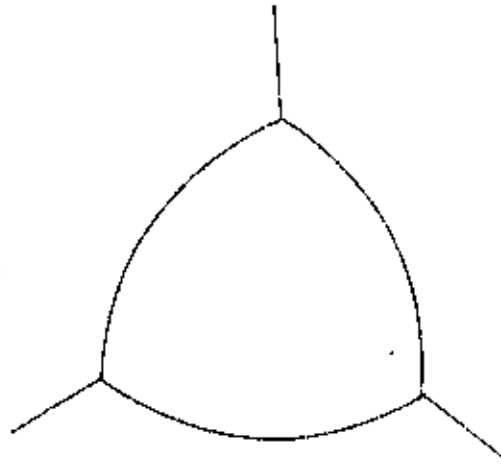


Fig.2.4 Sketch of a three-sided soap cell.

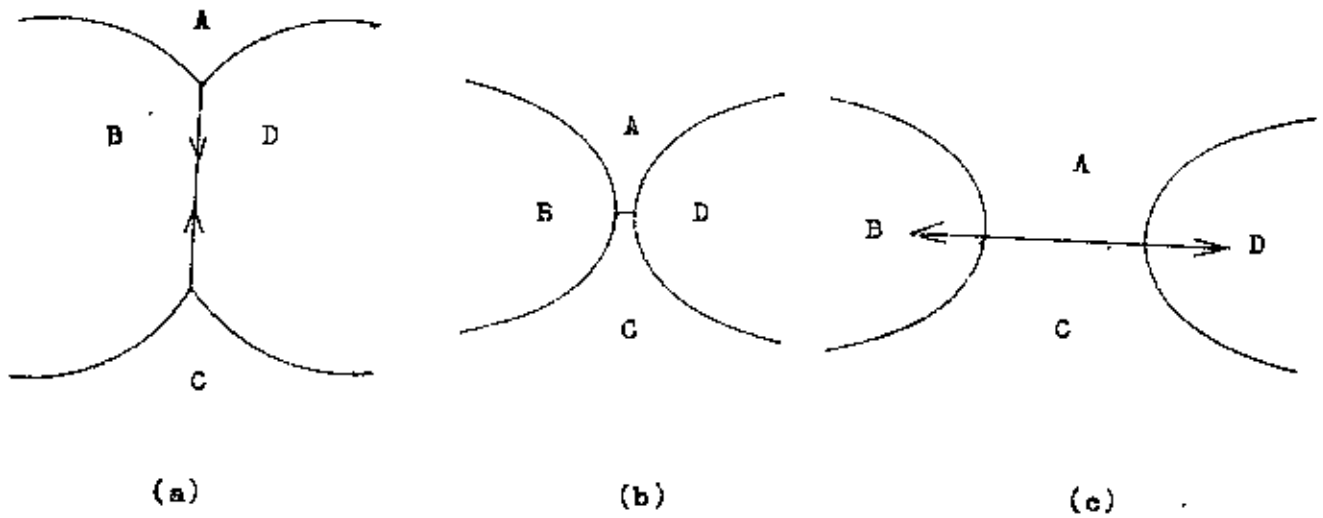


Fig.2.5 Sketch showing the mechanism that changes the number of sides of a grain during grain growth.

their centres, while those with more than six sides have walls convex toward their centres, this effect being more pronounced the greater the number of sides above six. This confirms the fact that the only geometrical figure formed by straight lines can have an average internal angle of 120° is the hexagon. In a two dimensional structure, such as that shown in the photographs of Fig. 2.3, all cells with less than six sides are basically unstable and tend to shrink in size, while those with more than six sides tend to grow in size as the walls tend to move towards their centre of curvature. Another fact is that there is a definite correspondence between the size of the cells and the number of the sides that they contain. The smaller cells usually have the fewer number of sides. It is no wonder that three sided cells disappear so rapidly, for both their small size and intrinsic number of sides require that their walls have very large curvatures, with accompanying high pressure differentials, diffusion rates, and rates of wall migration. The photographs show that, in general, four and five sided cells do not disappear as a unit, but first change to three-sided cells which rapidly disappear.

Another important aspect of cellular growth can be seen in the photographs of Fig. 2.3. This aspect has to do with the fact that over a period of time the number of sides that any given grain possesses continually changes. The number of sides may increase or decrease, as can be seen by considering the mechanism⁽²⁰⁾ illustrated in Fig. 2.5. Because of the curvature of the boundaries that separate cells B and D from A and C respectively, the boundaries migrate, thus eliminating the

boundary between cells B and D, and then create a new boundary between A and C. These steps are indicated in Fig. 2.8b and 2.8c respectively. As a consequence of this process, cells B and D each lose a side, while cells A and C each gain a side. Each time a three-sided cell disappears, each of its neighbouring cells loses one side.

It is interesting to observe in the pictures of Fig. 2.3 that, while the number of cells keep decreasing as time goes on, the cellular arrays are geometrically similar at all times. It is apparent that at any given instant the cells vary in size about a mean and that this mean size grows with time. The mean cell diameter serves as a convenient measure of the cell size of an aggregate. Therefore, when one refers to the cell size of a growth, it is the diameter of the average cell that is meant. This statement also holds true for metals where the commonly used term 'grain size' refers, in general, to the mean diameter of an aggregate of grains. It follows that grain growth, or cellular growth, refers to the growth of the average diameter of the aggregate.

2.2.2. Law Governing Grain Growth:

An expression for a soap froth that relates the size of the average cell to the time may be derived. It may be assumed that the rate of growth of the cells is proportional to the curvature of the cell walls of the average cell at a given instant of time. This assumption is in agreement with the fact that boundaries move as a result of glaucous diffusion caused by a pressure difference from one side of a soap film to the other, and with the fact that this pressure difference is proportional to the curvature of the boundary. If the symbol D

represents the mean diameter of the average-sized soap cell, and C the curvature of the walls of this cell, there may be an expression (21).

$$\frac{dD}{dt} = K'C$$

where ' t ' is the time, and K' a constant of proportionality. It may be also assumed that the curvature of the average sized cell is inversely proportional to its diameter, and rewrite the above equation in the following form :-

$$\frac{dD}{dt} = \frac{K}{D}$$

Where K is another constant of proportionality. Integration of this equation leads to the following result :

$$D^2 = Kt + C$$

Assuming that D_0 is the size of the average cell at the start of the observation ($t=0$), an evaluation of the constant of integration gives

$$D^2 - D_0^2 = Kt$$

Although several rather assumptions have been made in deriving the above equation, experimental measurements of the growth of cells in a soap froth have shown that this expression fits the observed data quite closely (21). It can therefore be concluded that the above equation is essentially correct for the growth of soap cells under the action of surface tension forces.

If it is assumed that the grain size is very small at the beginning of cell growth, then it is possible to neglect D_0^2 in relation to D^2 , with the result that the equation relating the cell size to the time can be expressed in the somewhat simpler form.

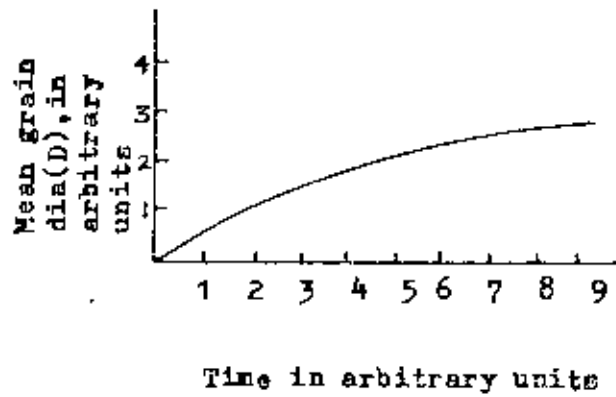


Fig.2.6 Graphical representation of the ideal grain growth law, $D = kt^{1/2}$

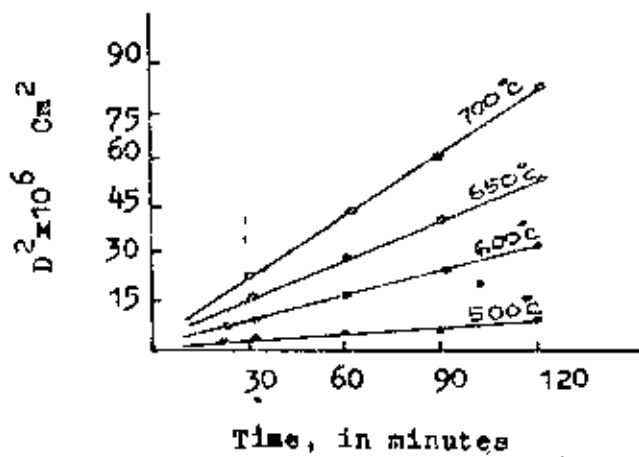


Fig.2.7 Grain growth isothermal for alpha-brass.

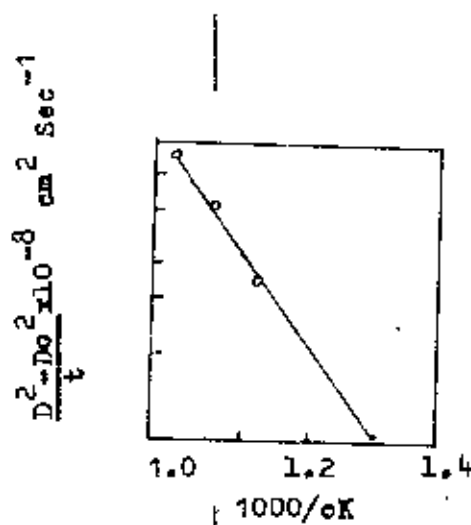


Fig.2.8 The logarithm of the slope of the isotherms of Fig.2.7 vary directly as the reciprocal of the absolute temperature

$$D^2 = kt$$

$$\text{or } D = kt^{1/2}$$

where $k = \sqrt{K}$. According to this relationship, the mean diameter of the cells in a soap froth grows as the square root of time. Fig. 2.6 shows a graph of this relationship, as time progresses; the rate of cellular growth diminishes.

Consider the case grain growth in metals may be considered. Here, as in soap froth, the driving force for the reaction lies in the surface energy of the grain boundaries. Grain boundary movements in metals are, in every respect, analogous to those of the cell walls of soap froth. In both cases, the boundaries move toward their centres of curvature, and the rate of this movement varies with the amount of curvature. On the other hand, while it is known that the growth of soap cells can be explained in terms of a simple diffusion of gas molecules by which atoms on one side of a grain boundary cross the boundary and join the crystal on the other side.

It has been proposed (22) and generally accepted that the boundary atoms in the crystal on the concave side of the boundary are more tightly bound than the boundary atoms in the crystal on the convex side, because they are more nearly surrounded by neighbouring atoms of the same crystal. This tighter binding of the atoms on the concave side of the boundary should have an effect on the rate at which atoms move across the boundary from one crystal to the other, since the rate of movement is greater from convex to concave crystals than vice-versa. The greater the curvature of the boundary, the greater should be this effect, and hence faster the movement of the crystal boundary. However,



because detailed knowledge of the structure of metallic grain boundaries is lacking, the exact nature of transfer mechanism by which atoms cross a boundary is still not known, and it is not possible to explain quantitatively the apparently irrational results obtained when studying the growth of crystals in a metal.

When a quantitative theory capable of explaining grain growth in pure metals and alloys is lacking, much is known about the cause of the apparent abnormal nature of metallic crystal growth. A number of these points will now be considered. If metallic grain growth is assumed to occur as a result of surface-energy considerations and the diffusion of atoms across a grain boundary, then it is to be expected that at any given constant temperature a grain growth law of the same form as that found in soap froth would be observed, namely,

$$D^2 - D_0^2 = Kt$$

Also, if the diffusion of atoms across a grain boundary is considered to be a simple activated process, then it can be shown that the constant K in the above equation can be replaced by the expression

$$K = K_0 e^{-Q/RT}$$

where Q is an empirical heat of activation for the process, T is the temperature in degree Kelvin, and R is the gas constant. The grain growth law can therefore be written as a function of both temperature and time in the following manner :

$$D^2 - D_0^2 = K_0 t e^{-Q/RT}$$

Most of the early experimental studies of metallic growth have failed to confirm the above relationship. However, certain work⁽²⁵⁾ has given results that confirm, quite closely, to the above equation. Fig. 2.7 shows part of these results for brass (10 percent zinc and 90 percent copper). In each case the straight lines are obtained when D^2 is plotted against the time. The grain-growth equation may be rearranged as follows:

$$\frac{D^2 - D_0^2}{t} = K e^{-Q/RT}$$

Taking logarithms of both sides of this equation, gives

$$\text{Log} \left(\frac{D^2 - D_0^2}{t} \right) = - \frac{Q}{2.3RT} + \text{Log } K$$

This relationship tells that the quantity $\log (D^2 - D_0^2)/t$ should vary directly as the reciprocal of the absolute temperature ($1/T$) and that the slope of this linear relationship is $Q/2.3R$. Now, the quantity $(D^2 - D_0^2)/t$ equals the slope of a grain growth isotherm, such as those shown in Fig. 2.7. Fig. 2.8 shows the logarithm of the slopes of the four lines in Fig. 2.7 ($\log (D^2 - D_0^2)/t$) plotted as functions of the reciprocal of the absolute temperature ($1/T$).

Most of the reported experimental work does not confirm so well to the grain growth equation. In order to compare these results with the grain-growth equation D_0^2 may be neglected when compared with D^2 , so that the ideal grain growth law can be stated in the simpler form.

$$D = kt^{1/2}$$

where the constant k equals \sqrt{K} and should be a function of temperature given below :

$$k = \sqrt{K} = \sqrt{K_0 \cdot e^{-Q/RT}} = k_0 \cdot e^{-Q/2RT}$$

where $k_0 = \sqrt{K_0}$

Many of the experimental isothermal grain-growth data correspond to empirical equation of the form

$$D = kt^n$$

where the exponent n is, in most cases, smaller than the value $\frac{1}{2}$ predicted by the grain-growth equation. Furthermore, it has been found that the exponent ' n ' is not constant for a given metal or alloy, if the isothermal reaction temperature is changed⁽¹⁸⁾.

Chapter -3

PLASTIC DEFORMATION OF METALS

3.1. Effects of grain size on yield point of metals.

In a polycrystalline metal the grains are forced to deform plastically on several independent glide systems⁽²⁴⁾. This is the main effect of polycrystallinity on the plastic properties of metals. In F.C.C. and B.C.C. metals it increases the rate of work hardening. In materials with fewer glide systems the effect is more drastic. The plastic strain is constrained to the same order as the elastic strain, until the applied stress is raised sufficiently high to bring the necessary additional glide systems into play.

The effect of grain size on the yield stress of steel is frequently given by the Hall⁽¹⁾-Petch⁽²⁾ equation

$$\sigma_y = \sigma_i + kd^{-1/2} \quad \text{-----(3.1)}$$

Where σ_y is the yield stress, d is the average grain diameter and σ_i and k are constants.

A slip band of length d (average diameter) across a grain can be considered; its dislocation piles up against the grain boundaries and produces a large concentration of stress in the nearby parts of the adjoining grains (Fig. 3.1). If the applied stress is σ and the friction stress which opposes the glide of a dislocation in a slip band is σ_f (where $\sigma > \sigma_f$) then, when the slip band forms, a

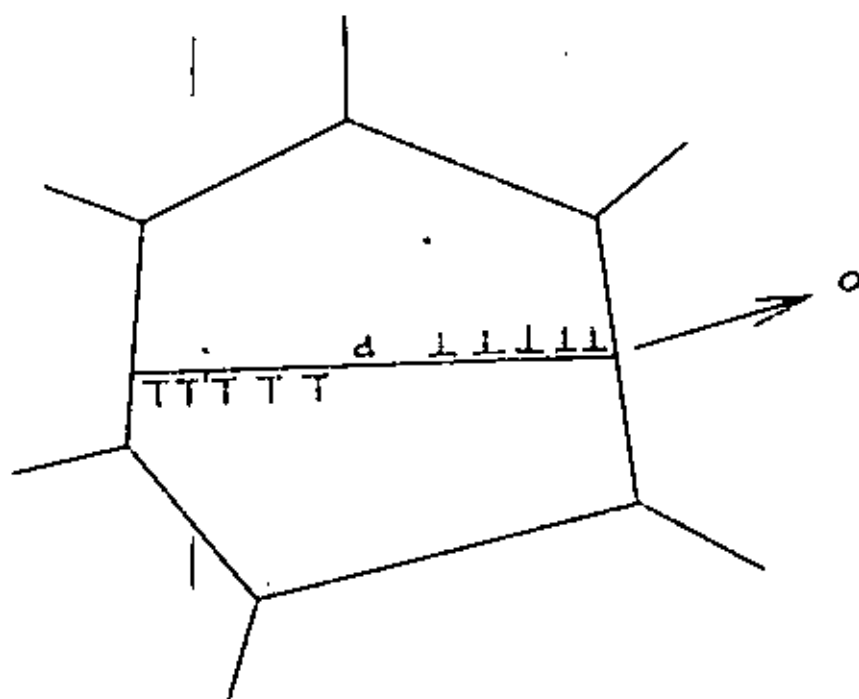


Fig.3.1 Pile-ups of dislocations against grain boundaries :
at ends of a slip plane.



where $\sigma_y = \sigma$ and $k = (\sigma - \sigma_1) / \sigma_1$

which is equivalent to Hill-Beck formula

$$\sigma = \sigma_1 + \sigma_2 P^{1/2}$$

yield is $\sigma_1 = (\sigma - \sigma_1) \times (P/n)^{1/2}$ or at an applied stress

be σ_1 . Then the pile-up stresses can make the neighbouring grains elastic, but the local stress needed to trigger yielding at any point

effect of a foreign inclusion is it, but its neighbours are still

2.1. In fact so yield (e.g. a source of the stress concentrating

yield point before it can become plastic. The central grain in fig.

only pinned by impurity atoms there is an stress up to its upper

Grain is an very pile-up. In metals in which the dislocations are st-

tronal local is thrown on to them and there is then little effect of

them, the adjoining grains begin to yield almost as soon as some addi-

dislocations which may have in many differently oriented glide sys-

gle up stresses, but the cracking may occur. In metals with free

the neighbouring grains may be uniaxially oriented for relaxing the

There can be several effects of this. In a material where

ring grain can not usually withstand it.

where $r \ll d$ this is a high concentration of stress and the neighbour-

is an additional stress of about $(\sigma - \sigma_1) \times (d/r)^{1/2}$. It points

length e. Similarly, at the distance r ahead of the slip band there

by a factor of about $(d/r)^{1/2}$ at a distance r ahead of a crack of

from elasticity theory that the applied stress is locally increased

directly similar to the stress concentrating near a crack. It is known

the material just beyond the ends of the slip bands. This effect is

load $(\sigma - \sigma_1) \times d$ is transferred from it (per unit thickness) on to

where impurity-pinning is weak, it seems that σ does not represent the stress to unpin dislocations in the adjoining grains. This is because the concentration of stress falls off as $r^{-3/2}$ from the end of the pile-up. To produce a moderate increase of stress in the nearest pinned dislocation, which may be some 10^{-4} cm from the pile-up, a concentration is required in the grain boundary at the tip of the pile-up which may exceed the ideal strength of the material there. In strongly pinned metals it is then easier to propagate Luders bands by creating dislocations from the grain boundaries where the pile-ups occur rather than by unpinning dislocations in the next grains. The same argument shows that even in weakly pinned metals, where unpinning is the easier process, a sizeable $\Delta\sigma$ is to be expected when the $r^{3/2}$ factor in $\sigma_i^{3/2}$ is large. In fact even the F.C.C. metals, which do not usually give marked yield-drops, generally follow equation 3.1 with an appreciable variation of yield stress with grain size.

3.2 Fracture Mechanics and Grain Size:

The phenomenon of the fracture of metals is complicated. Specimens fractured in tension vary greatly in appearance. Under tensile load a metallic material of extreme ductility may form a neck and separate without actual fracture. However, if the material can resist the tendency to neck sufficient enough to support transverse tensile stress a complex tensile stress is formed at the center of the test bar. The complex tension subjects the specimen to a smaller shear stress $(\sigma_1 - \sigma_2)/2$, instead of the shear stress produced by the axial tension

σ_f (Fig. 2.10). At the same time the ability of the specimen to deform also is decreased because the metal is less able to contract laterally as it is elongated axially. A fracture takes place first at the center of the necked region. As the strain increases the crack opens up gradually, roughly at right angles to the tensile force. Ultimately, the solid metal remaining around the periphery of the specimen fails in shear giving cup and cone fracture resembling ductile fracture. In some instances, by the time the central crack propagates enough to free the peripheral metal from its restraints, the load may be so great or the ability of the peripheral metal to deform plastically may be so small, that it fails immediately without stretching at all. Such a fracture may show little or no transverse contraction and is generally classed a brittle fracture through it is of the low ductility shear type. If the test specimen has a low ability to deform, the final fracture of the peripheral material might be started by small surface defects in the material which cause it to tear.

Thus fracture process generally requires two stages, i) creation of a crack, ii) growth of the crack. A very long pile-up of dislocations driven against an unyielding obstacle might lead to a crack through the coalescence of the leading dislocations (Fig.2.11) or the stress at the head of the pile-up may exceed the theoretical strength of the metal⁽¹⁸⁾. Naturally the growth of the crack formed is the critical stage. When the yield stress exceeds the growth stress of the cracks produced by the yield process, the material is brittle. If the converse is true small microcracks may form at the yield point but these will

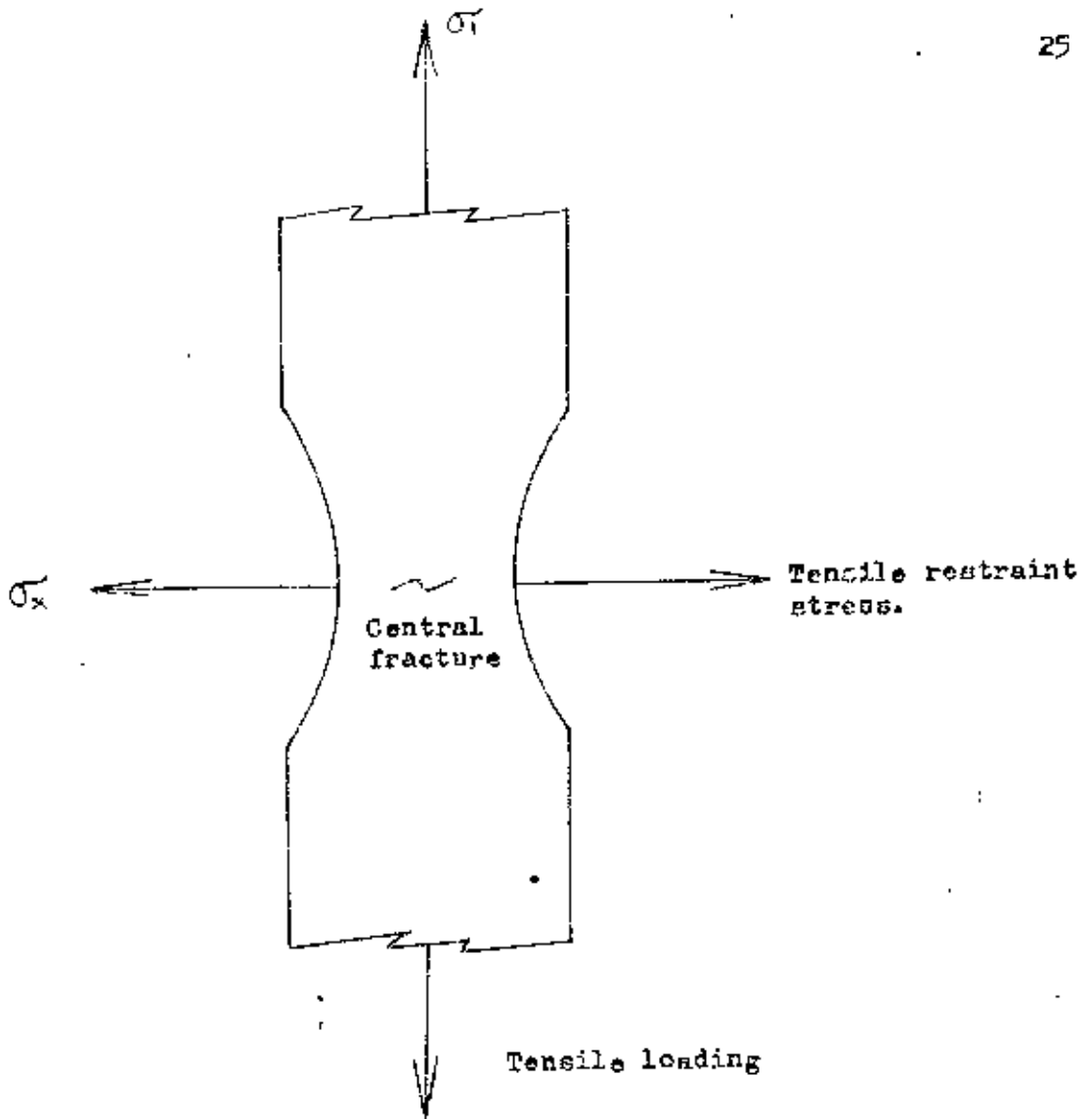


Fig.3.2 Formation of a crack due to peripheral stress.

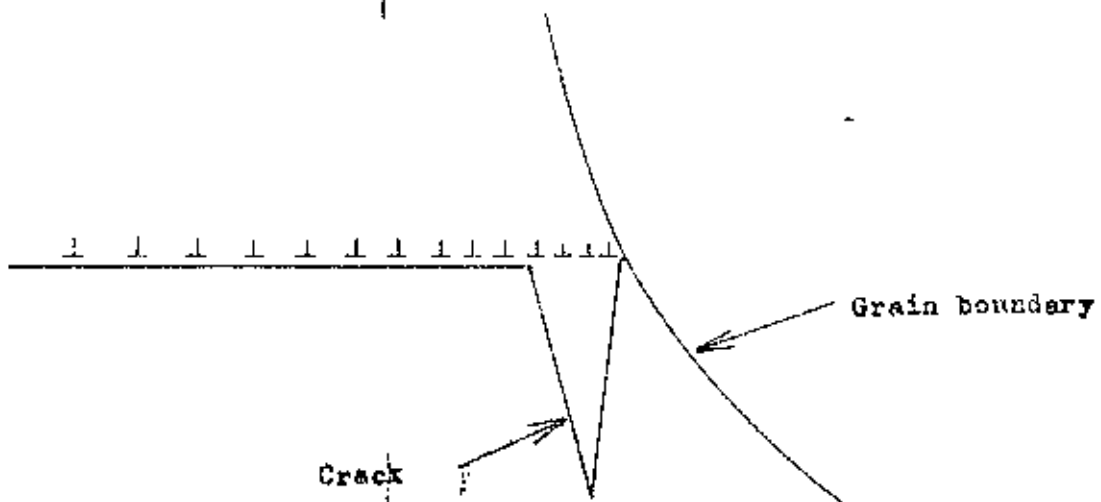


Fig.3.3, Nucleation of a crack by the coalescence of dislocations.

be ductile.

For a crack to extend to catastrophic dimensions the metal must have work-hardened to a high stress level. If by work hardening, the stress is raised to such a value that the Griffith⁽²⁶⁾ stress equation,

$$\sigma = \left(\frac{2E\gamma}{\pi C} \right)^{1/2}$$

Where E is Young's modulus
 γ is surface energy
 C is crack length.

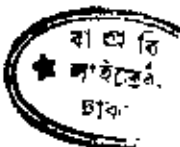
is satisfied for any crack which may have been produced the fracture of the specimen is a real possibility.

The fracture stress of polycrystalline metals usually exhibit the same form of dependence on grain size as does the yield stress. Cottrell⁽²⁷⁾ developed a theory which explains why the fracture stress varies with the inverse square root of the grain size. As regard grain size, small grains produce short shear bands and hence, short intercracks, large grains on the other hand produce long microcracks. Thus a Griffith like relation⁽²⁵⁾ can be expected

$$\sigma_f = k_f d^{-1/2}$$

with the grain size d taking place of c . Since $\sigma_y = \sigma_0 + k_y d^{-1/2}$

there may have a grain size transition if $k_f > k_y$; then $\sigma_f > \sigma_y$ when the grains small, and $\sigma_f < \sigma_y$ when coarse. Thus a ductile brittle transition at constant temperature due to the variation of grain size may be expected.



Chapter 4

EXPERIMENTAL PROCEDURE AND RESULTS

4.1. Production of Austenitic Grain Size.

In as much as austenite normally exists above the thermal critical range, it is of practical importance that the steel in question be so treated that the outlines of the prior austenitic grains will be apparent at room temperature upon subsequent microscopic examination. The treatment to which the steel is subjected for this purpose may produce a change in the chemical composition of the specimen, such as is brought about by oxidation and carburization methods; or the method may be one in which the composition of the specimen remains substantially unaltered. In most cases the latter method is preferred. Fortunately, by an appropriate cooling schedule from the significant heat treating temperature, the austenite may be transformed in such a manner that the prior austenitic grain boundaries will be clearly revealed. This may be brought about in one of the following ways⁽²⁸⁾:

1. By the rejection of either ferrite or iron carbide into the austenitic grain boundaries, forming a complete or nearly complete network, in hypoeutectoid and hypereutectoid steels, respectively.
2. By the formation of a transformation-product network at the prior austenitic grain boundaries as secured by appropriate interruption of the austenite transformation.
3. By the etch contrast, established between differently oriented grains of martensite, which have limits of orientation related to that of the parent austenite.

4. By high-temperature etching methods.

The first method has been used for the production of grain size. Plain carbon steels with a carbon content remote from the eutectoid composition will reject, upon slowly cooling through the transformation range, either ferrite or cementite, depending upon the carbon content of the steel. The rejection of these constituents occurs largely at the austenitic grain boundaries and, under appropriate conditions, the constituents may build up a complete or nearly complete network around the original grains, after which the remaining austenite transforms into pearlite or spherulite depending upon the rate of cooling.

4.2. Measurement of Grain Size:

The effect of grain size on mechanical properties immediately interests the engineer in establishing parameters for the size of the grains. In the methods to be discussed for rating the grain size of a metal or an alloy, estimating is facilitated by projecting the image of the structure on the focusing screen of a metallograph, or in some cases by obtaining a photomicrograph of a structure at some predetermined magnification. The prepared metallographic specimen in most cases must be suitably etched to produce good visual contrast between the grains and the grain boundary constituent.

The grain size may be reported in a number of ways, the following designations being acceptable, and the first two being most common.

By trial and error a match is secured and the grain size of the specimen is then appropriately designated by a number of the matching chart. Specimens showing a wide grain size are noted in a station number, and at least every 100 specimens a report the grain size in terms of two numbers, designating the approximate percentage of each size present.

pared with a series of graded standard grain size charts. A photomicrograph of the structure at the same magnification, is compared with the standard specimen project at a magnification of 100 X, or one of comparison, in which the field of the structure of an upper the basic indexing principle is involved. The method is essentially the same as the preferred method in most cases, owing to the high reliability of the procedure, and the widespread acceptance of grain size as the preferred method of designating the grain size.

1. ASTM (American Society for Testing Materials) Grain size number - arbitrary expansion numbers with reference to the number of grains per square inch at a magnification of 100 X.
 2. The average number of grains intercepted by a line of a given diameter length (Grain Cylinder Intercept with d).
 3. The average number of grain per square millimeter.
 4. The average area of grains expressed in square millimeters.
- The first method was used for the measurement of grain size.

The ASTM austenitic-grain-size standards have been designed to cover the range of grain sizes normally encountered in most grades of steel. The standard charts are indexed from No. 1 to No. 8, each index number representing some mean number of grains per square inch at a magnification of 100 X according to the following relationship⁽³¹⁾.

$$\text{Mean number of grains per square inch at 100 X} = 2^{n-1}$$

where n = ASTM grain-size index number.

Data concerning the ASTM grain size standards and the interrelationship between index number, mean number of grains per square inch at 100 X, and the actual existing grain size as determined by calculations, are shown in Table 4-1 XXXXXXXXXX

For all practical purposes, grain size Nos. 1 to 5 may be considered to comprise the coarse-grained steels, whereas Nos. 5 to 8 may be considered as fine grain steel. Size No. 5 may be considered as either coarse or fine, depending upon whether or not those relatively few grains which are outside the No. 5 range are coarser or finer than No. 5.

It is not uncommon to encounter austenitic grain size at 100 X that are either larger than ASTM No. 1 or smaller than No. 8. Larger grain sizes can be appropriately rated by using a projected magnification of 50 x, instead of 100 x and securing a match with one of the standard charts in the usual manner. At this lower magnification, however,

Table 4-1

ASTM grain- size number	No. of grains per sq. in at 100 X		Calculated diameter of equivalent spherical grain, not magnified	
	Mean	Range	In	Mil.
1	1	0.75-1.5	0.01130	0.287
2	2	1.5 -3	0.0080	0.203
3	4	3-6	0.00567	0.144
4	8	6 -12	0.00400	0.101
5	16	12 - 24	0.00283	0.0718
6	32	24- 48	0.00200	0.0507
7	64	48 - 96	0.00142	0.0359
8	128	96 - 192	0.00100	0.0254

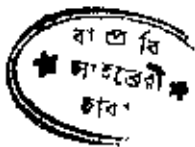
grain sizes Nos. 00 and 0 are reported when matched to standard charts Nos. 1 and 2, respectively. In the case of grain sizes smaller than No. 8, a magnification of 200 X may be used, and grain sizes No. 9 and 10 reported when rated against standard Nos 7 and 8 respectively.

4.3. Material Used and Results Obtained:

The material used in the experiment was mild steel containing approximately 0.35% carbon. The microstructure of the mild steel is shown in Fig. 4.1.



Fig. 4.1 Microstructure of the mild steel used in experiment.



The as received material was hot-rolled stock in the form of 1.5 inch diameter round bar. Sections of 10 inch long were cut from the bar and heat-treated in a silicon carbide heating furnace at different temperatures to have different grain sizes. Each section was heated slowly upto the desired annealing temperature, kept at that constant temperature for 2 hours and then slowly cooled in the furnace just below the upper critical (A_2) temperature (Fig.4.2). At this temperature the specimen was kept for about 15 minutes* (Fig.4.3), so that proeutectoid ferrite can come out from austenite and take their place along the grain boundaries. Without this treatment it is difficult to observe the grain boundary. The specimen was quenched in water so that no further change in structure can take place.

Then a 1/2 inch section was cut from the original heat-treated 10 inch long rod. That piece was properly polished, etched and then viewed under a metallurgical microscope. Then the grain size was determined by ASTM method. Data of the experiments performed to obtain different grain sizes is given in table 4.2. The idealised, standard ASTM grain sizes and the grain sizes observed under microscope for various heat-treated specimens are matched in Figs. 4.4 to 4.11.

Each of the quenched specimen was tested to determine its hardness. Table 4-3 shows the annealing temperature quenching medium used, and the hardness obtained due to the quenching.

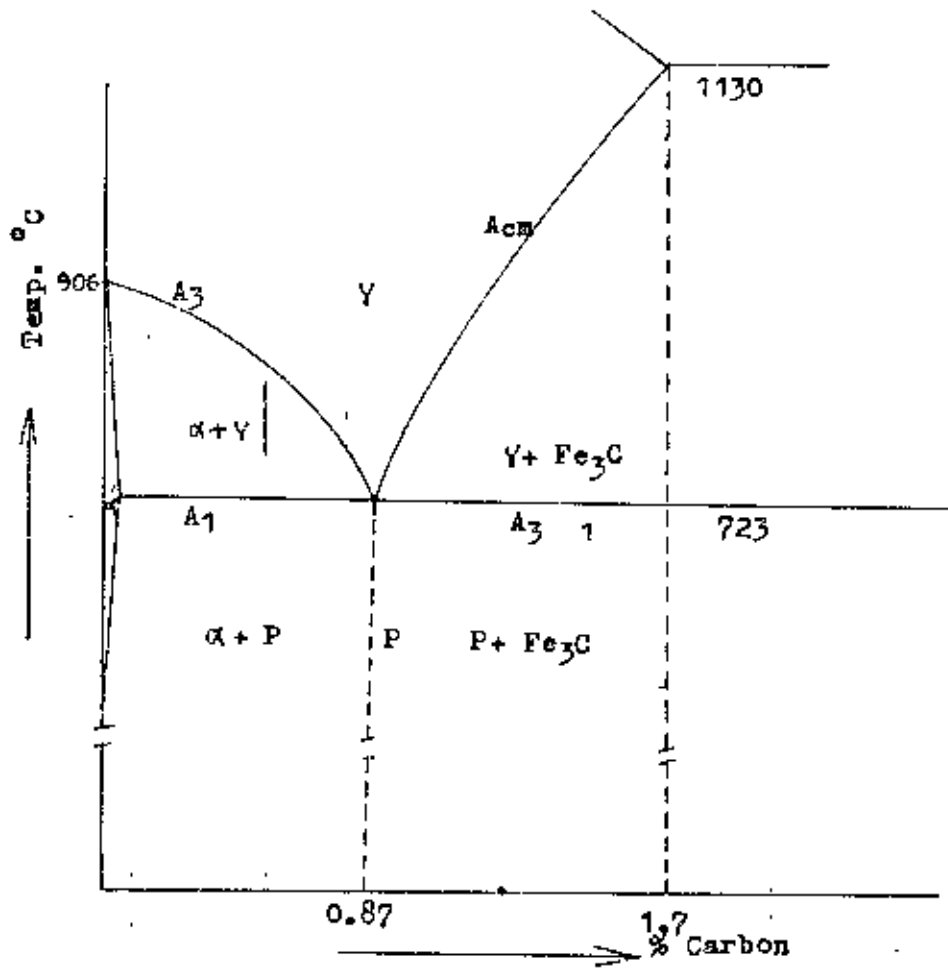


Fig.4.2 Iron carbon diagram showing different phases during cooling.

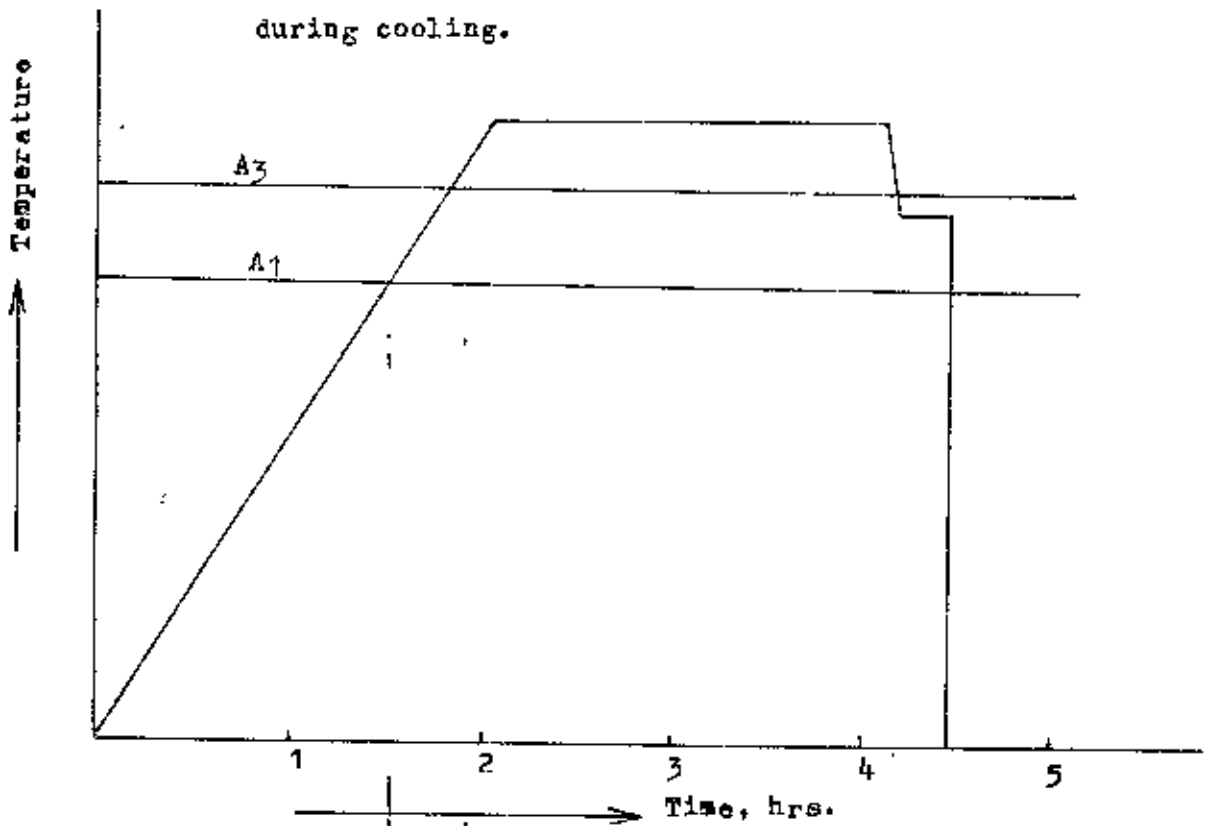


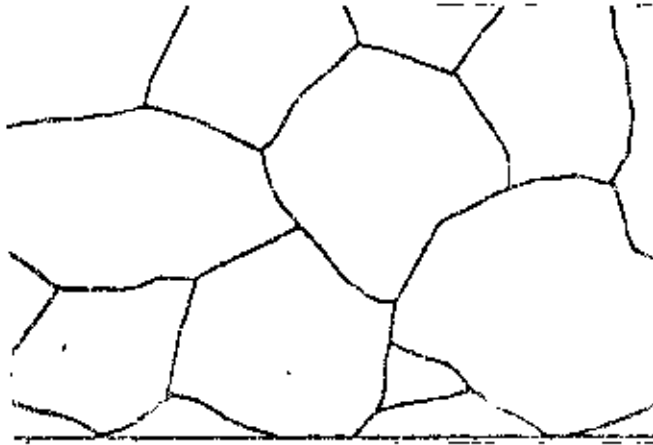
Fig.4.3. Heat treatment cycle of the specimens.

Then tensile test specimens were machined from each quench of 10 inch long original bar. The rest portion of the bar was then tempered at a temperature of 600°C in a furnace for $2\frac{1}{2}$ hours. Tensile test specimens were also machined from each of the tempered samples. Then both the quenched and tempered specimens were tested to determine the yield strength and ultimate strength. The test results are given in table 4.4 and 4.5.

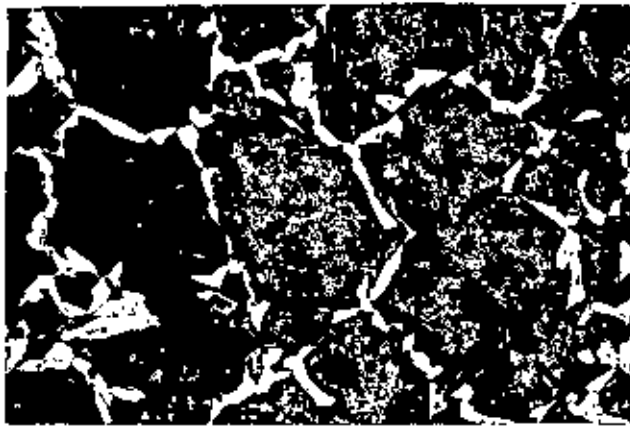
Table 4-2

Specimen heat-treatment

Specimen Condition	Annealing Treatment	Post transformation annealing treatment	Quenching Medium	Average grain size according to ASTM method
1	2 hr. at 1150°C	15 min. at 760°C	Water	1
2	2 hr. at 1100°C	15 min. at 760°C	Water	2
3	2 hr. at 1025°C	15 min. at 770°C	Water	3
4	2 hr. at 1000°C	15 min. at 770°C	Water	4
5	2 hr. at 975°C	15 min. at 770°C	Water	5
6	2 hr. at 950°C	15 min. at 770°C	Water	6
7	2 hr. at 900°C	15 min. at 770°C	Water	7
8	2 hr. at 870°C	15 min. at 770°C	Water	8



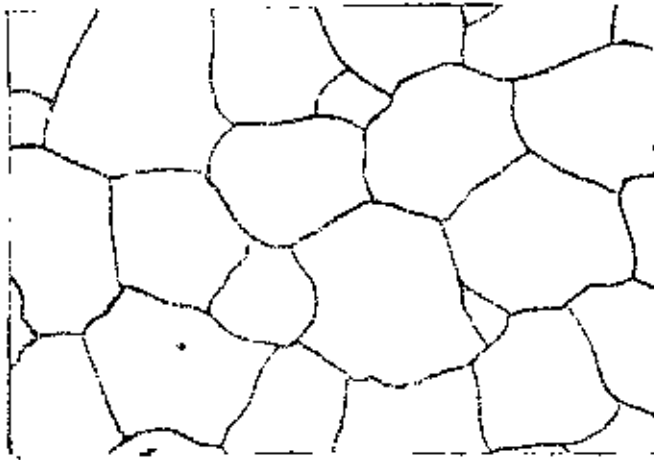
(a)



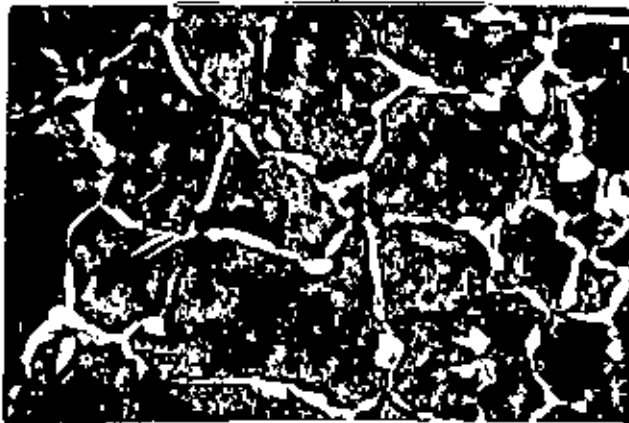
(b)



Fig.4.4 (a), ASTM, standard grain size No.1⁽³⁰⁾. Up to 1½ gr. per sq.in. at 100X. (b) observed microstructure of specimen No.1 (Table 2) which is equivalent to grain size No.1 at 100X.

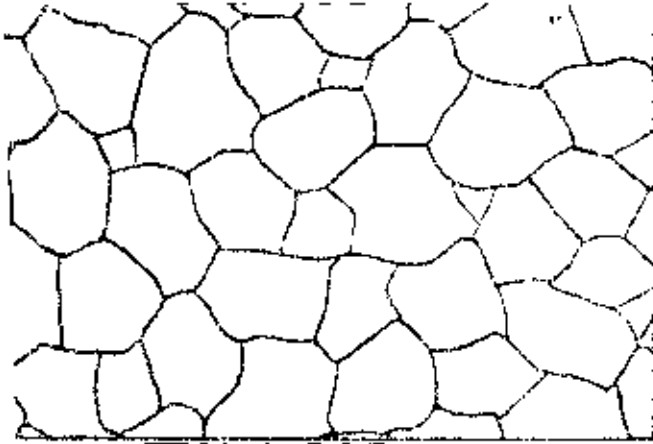


(a)

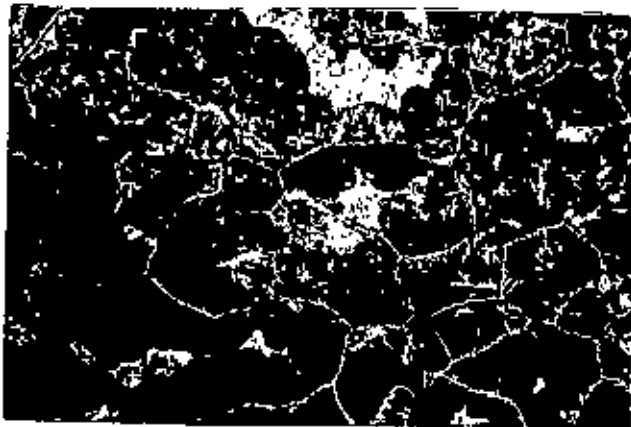


(b)

Fig. 4.5(a) ASTM standard grain size No. 2. $1\frac{1}{2}$ to 3 gr. per sq. in. at 100X, (b) observed microstructure of specimen No. 2 (Table 2) which is equivalent to grain size No. 1 at 100X.

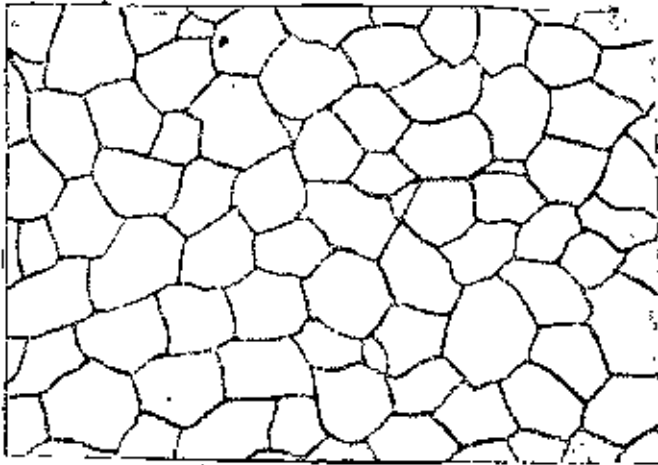


(a)



(b)

Fig.4.6(a), ASTM standard grain size No.3.3 to 6 gr. per sq. in at 100X,(b)observed microstructure of specimen No.3(Table-2) which is equivalent to grain size No.3 at 100X.

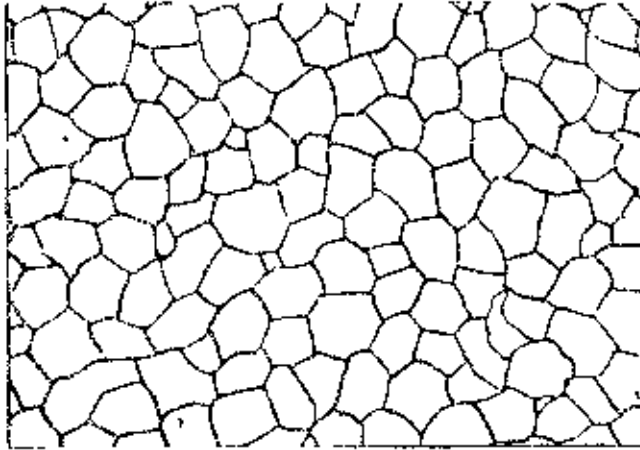


(a)

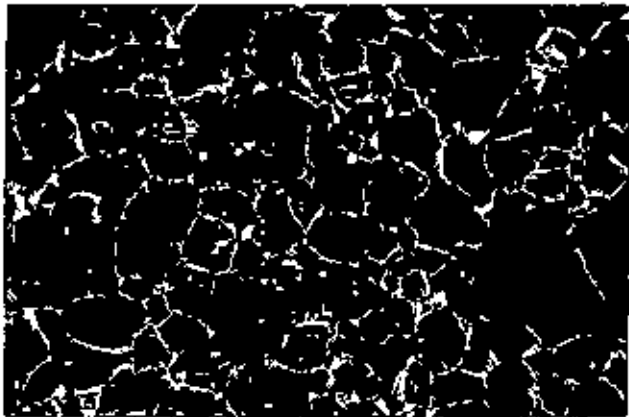


(b)

Fig.4.7 (a) ASTM standard grain size No.4. 6 to 12 gr. per sq. in. at 100X, (b) observed microstructure of specimen No.4 (Table 2) which is equivalent to grain size No.4 at 100X.

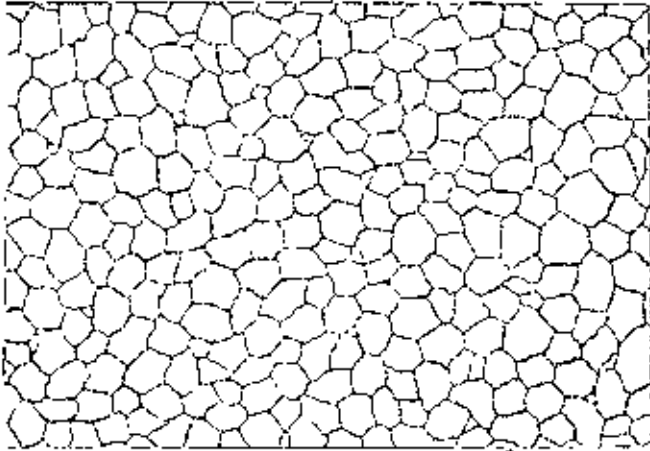


(a)

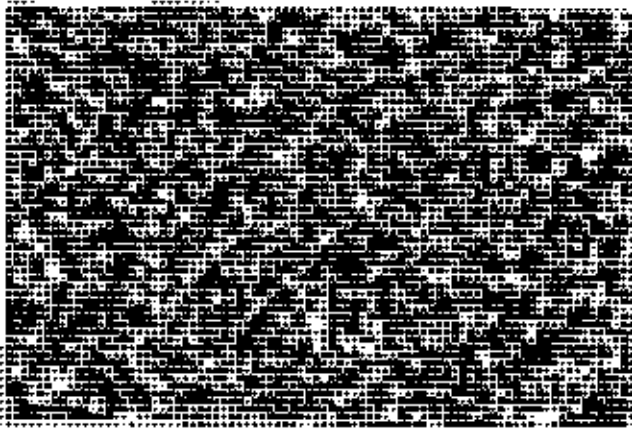


(b)

Fig.4.8 (a) ASTM standard grain size No.5. 12 to 24 gr. per sq. in. at 100X, (b) observed microstructure of specimen No.5 (Table 2) which is equivalent to grain size No.5 at 100X.

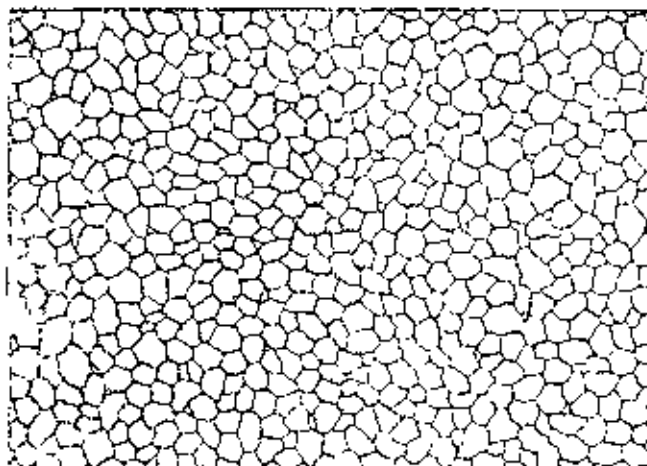


(a)

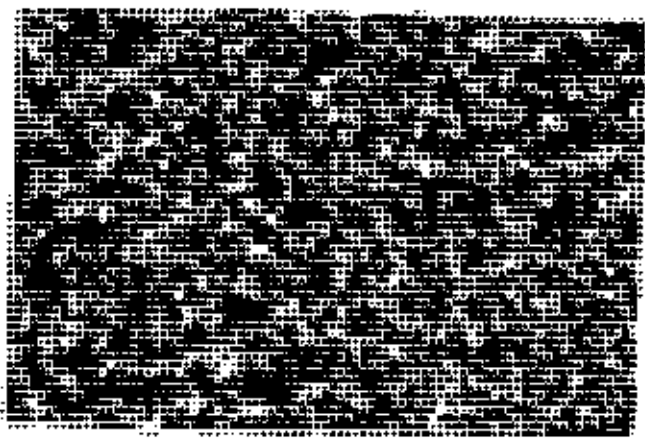


(b)

Fig. 4.9 (a) ASTM standard grain size No. 6. 24 to 48 gr. per sq. in. at 100X (b) observed microstructure of specimen No. 6 (Table 2) which is equivalent to grain size No. 6 at 100X.



(a)



(b)

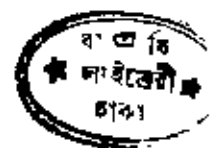
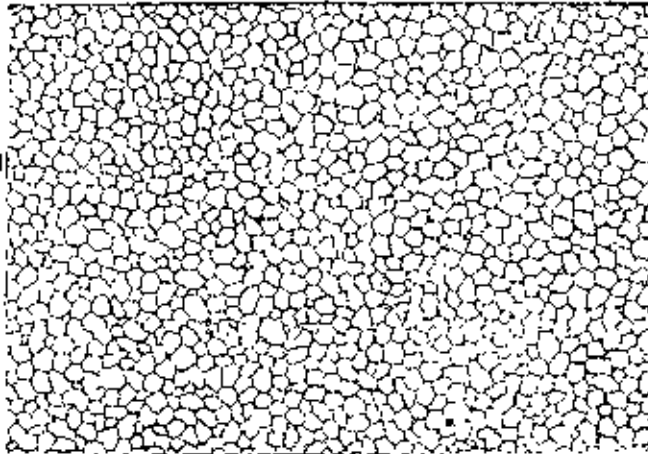


Fig.4.10 (a) ASTM standard grain size No.7. 48 to 96 gr. per sq. in. at 100X. (b) observed microstructure of specimen No.7 (Table 2) which is equivalent to grain size No.7 at 100X.



(a)



(b)

Fig. 4.11(a) ASTM standard grain size No. 8. 96 to 192 gr per sq. in. at 100X, (b) observed microstructure of specimen No. 8 (Table 2) which is equivalent to grain size No. 8 at 100X.

Table 4.3

Hardness of specimens after quenching (as quenched condition).

Specimen No.	Annealing Treatment	Quenching Medium	Average grain size according to ASTM method	Hardness Rockwell-C Scale Load 150 Kg.
1	2 hr. at 1150°C	Water	1	31.8
2	2 hr. at 1100°C	Water	2	27.7
3	2 hr. at 1025°C	Water	3	24.3
4	2 hr. at 1000°C	Water	4	24.0
5	2 hr. at 975°C	Water	5	23.9
6	2 hr. at 950°C	Water	6	20.2
7	2 hr. at 900°C	Water	7	20.1
8	2 hr. at 870°C	Water	8	14.3

Table 4.4: Results of tensile tests (as quenched condition).

Sample No.	Grain Size No.	Average Grain Diameter in.	(Grain Dia) ³ d ³	Total body-dry area per cubic inch of steel(2) in ²	Yield load lb.	Yield strength x10 ³ PSI	Percentage elongation 2" gauge length
1	1	0.01130	9.40	300	8,800	44	2.5
2	2	0.00800	11.18	424.3	9,400	47	5.5
3	3	0.00567	13.28	600	10,400	52	1.25
4	4	0.00400	15.81	648.5	11,400	57	3
5	5	0.00283	18.80	1200	13,000	65	3
6	6	0.00200	22.36	1697	14,000	70	7.25
7	7	0.00142	26.54	2400	15,400	77	11
8	8	0.00100	31.62	3394	17,600	88	4

Table 4.5

Results of tensile tests (after tempering at 600°C for 2½ hrs.)

Sample No.	Grain size No.	Average grain Diameter in.	(Grain Dia) - $\frac{1}{2}$ $\frac{1}{2}$	Total boundary area per cubic inch of steel(2) in ²	Yield load lb.	Yield Strength $\times 10^3$ PSI
1	1	0.01130	9.40	300	0.400	42
2	2	0.00800	11.18	424.3	9.000	45
3	3	0.00567	13.28	600	10.000	50
4	4	0.00400	15.81	848.5	10.800	54
5	5	0.00283	18.80	1200	12.400	62
6	6	0.00200	22.36	1697	13.600	68
7	7	0.00142	26.54	2400	15.200	76
8	8	0.00100	31.62	3394	17.000	85

Chapter 5

DISCUSSION AND CONCLUSION

The results of table Nos. 4.2- 4.5 are represented graphically in Fig. Nos. 5-1 - 5.3. The relation between the annealing temperature and the grain sizes are shown in Fig.5-1. The annealing temperature for each case was 2 hrs. The curve shows that very fine grains are less prone to growth with higher temperature, because diffusion is less at lower temperatures. On the other hand very big grain also change slowly with temperature because there are few small grains at whose expense they can grow⁽³²⁾. Besides they may have more six sided grains which is rather stable (see sec. 2.2.1). Intermediate size grains are more sensitive to temperature fluctuation, because less than 6 sided cells are present there and the temperature is sufficient to cause diffusion at a faster rate which results in the migration of grain boundaries.

Fig. 5.2 shows the relation between the grain size and hardness obtained. The samples and hardness obtained. The samples and hardness obtained. The samples were not quenched directly from the annealing temperature. Rather, they were cooled from annealing temperature to just below the upper critical (A_3) temperature and kept there for 15 minutes (Fig.4.3) to give sufficient time so that proeutectoid ferrites come out from the austenite and take their place along grain boundaries. The grain size obtained at higher temperature is unaffected by this cooling unless the time is increased very considerably.

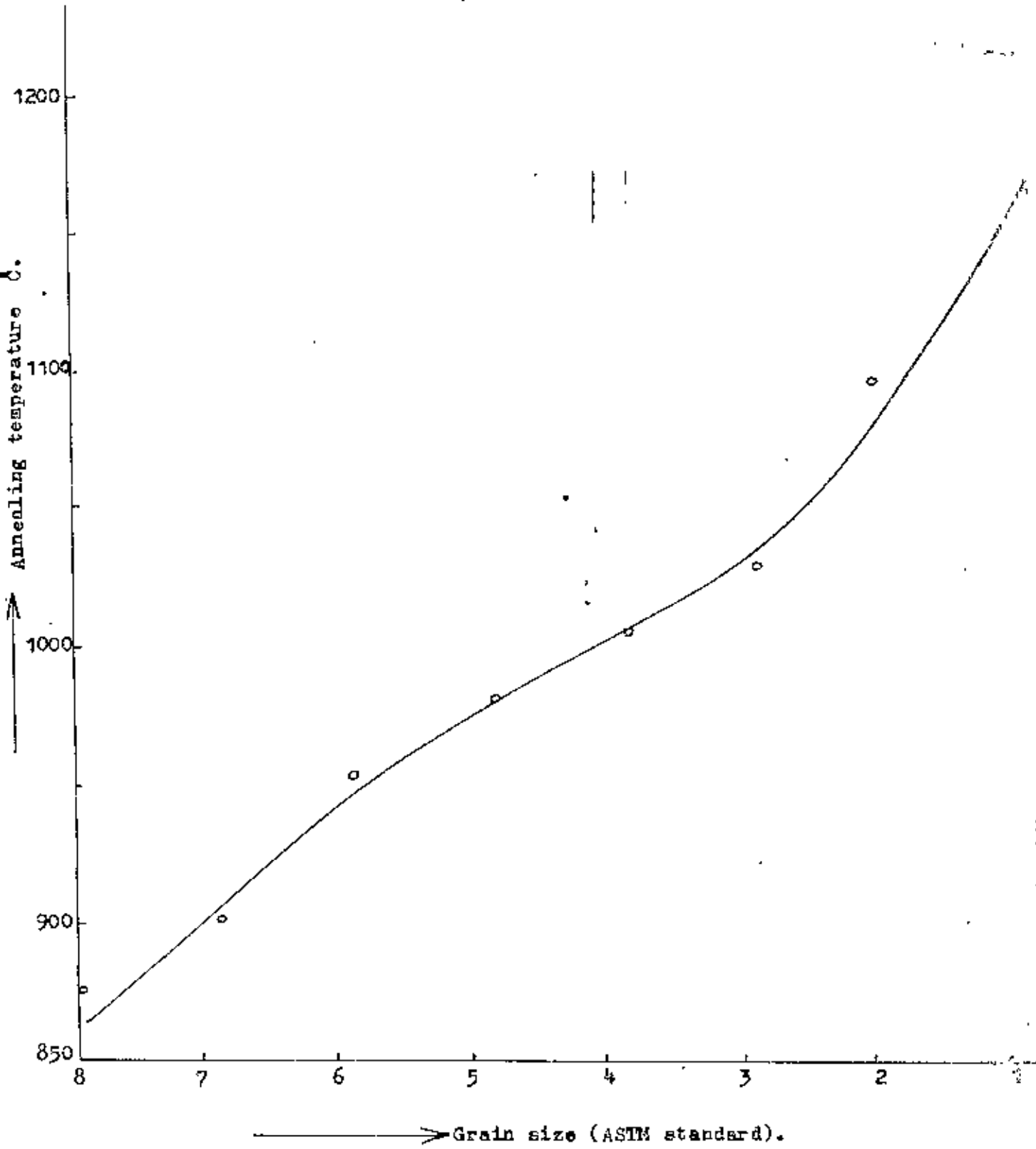


Fig.5.1 Effect of annealing temperature on grain size.

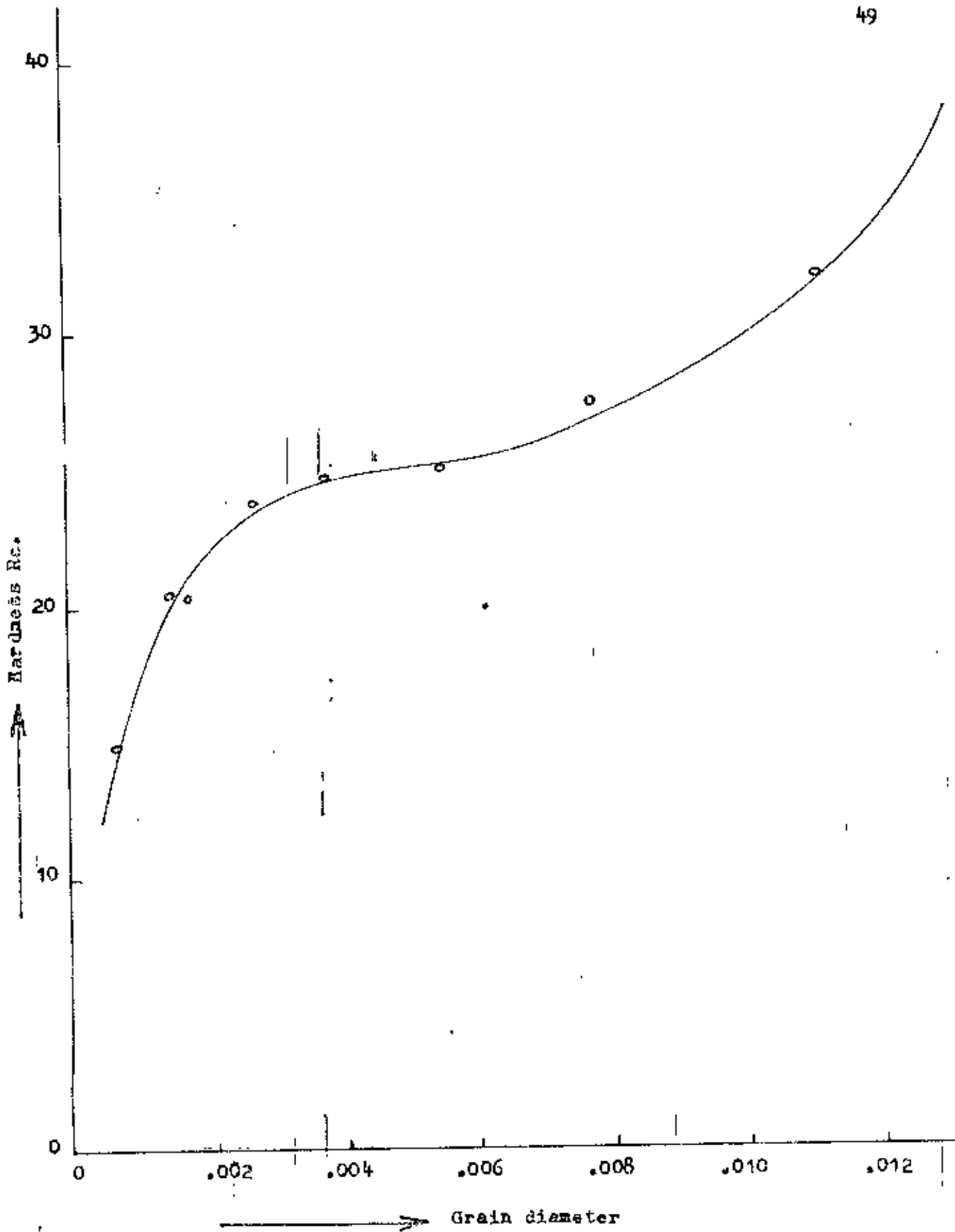


Fig. 5.2 Hardness Vs. Grain diameter (as quenched condition).

The samples were then quenched from this post annealing temperature. The curve shows that with increasing grain size the hardness is also increasing, which is contrary to the normal structure where hardness should fall with increase in grain size. This anomaly may be explained in the following manner. Rate of cooling was not equilibrium. Stored energy were higher in specimens having annealing temperatures when they were cooled to just below upper critical temperatures. As a result specimens with higher annealing temperatures were more affected by quenching and the stress produced were more, resulting in higher hardness.

Fig. 5.3 shows the relation between the grain size and yield strength of as quenched and tempered specimens. The tempered specimens. The tempered specimens were heat-treated at a temperature of 600°C for $2\frac{1}{2}$ hours since there is no significant fall of hardness with further rise in temperature (see Appendix I). Both the curves show that the yield strength increase with the decrease of grain size and it varies linearly with $d^{-\frac{1}{2}}$. This result confirms Hall-Petch equation. The increase in the yield strength is due to the pile-up of dislocations at the grain boundary which acts as a barrier. For smaller grains the number of grain boundary intercepts are more and so the yield strength is higher. Fig. 5.4 shows that yield strength increases at a faster rate in a similar manner as the grain boundary area increases with the decrease of grain diameter per unit volume of the specimen. This is analogous to the conclusion drawn from fig. 5.3. Again a study of Fig. 5.3 suggests that tempering may not have any effect on yield strength.

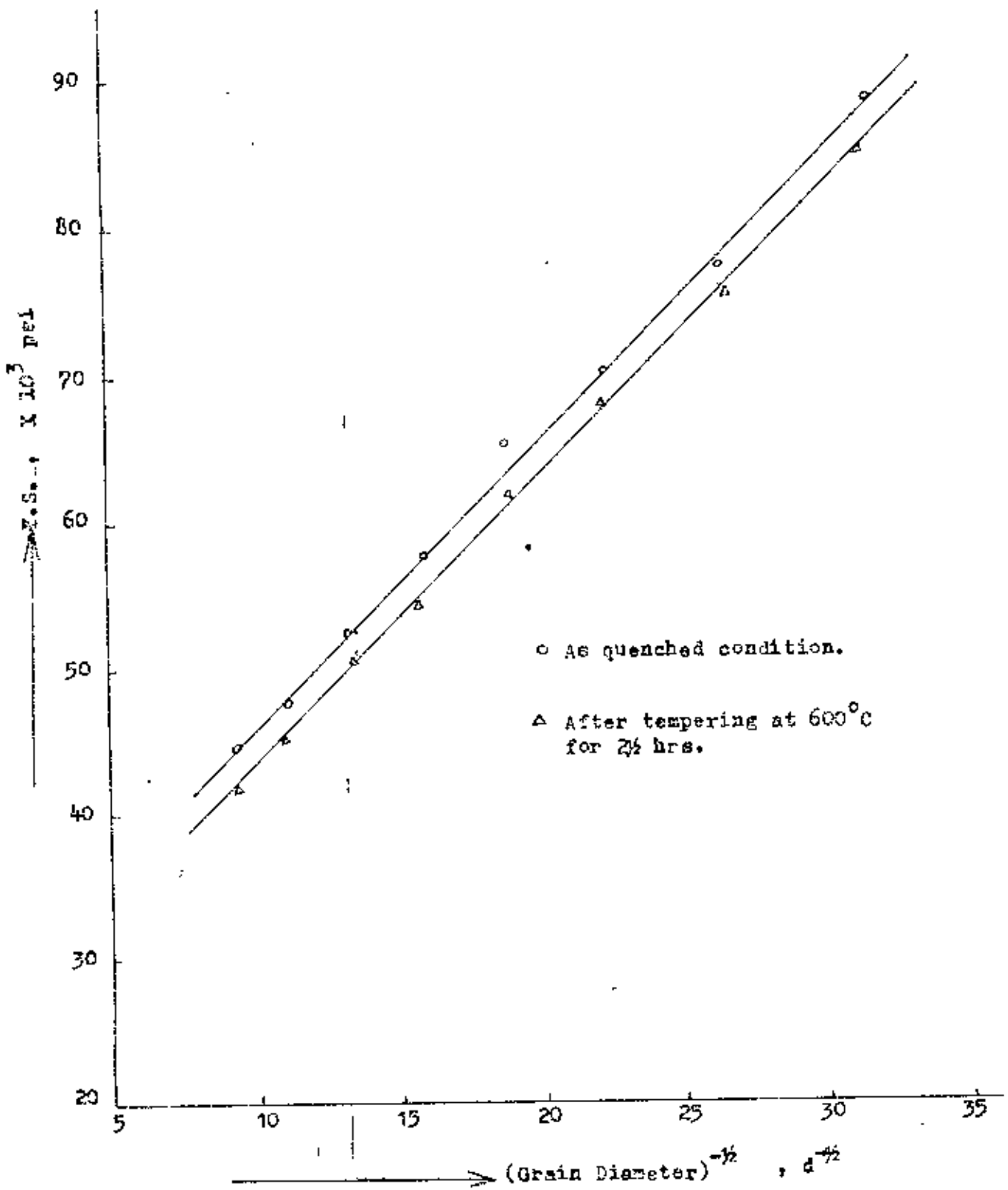


Fig.5.3 Effect of grain size on tensile strength.

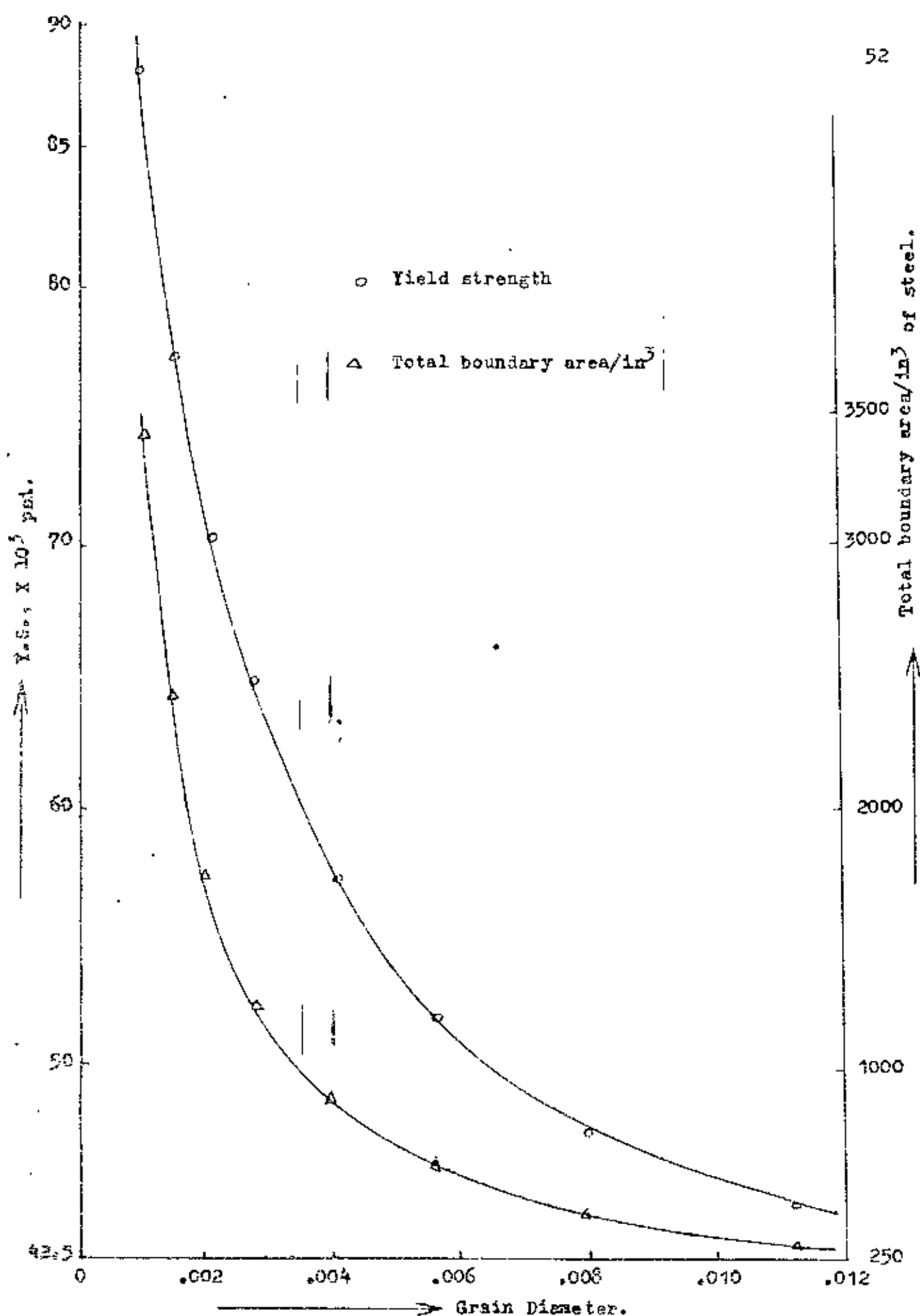


Fig. .4 Effect of grain size on total boundary area per cubic inch of steel and tensile strength (as quenched condition).

In Fig.5.5 the percentage elongation is plotted against $d^{-1/2}$. The stress at which a ductile specimen fractures is not an especially meaningful physical quantity since it depends on the surface of the specimen. However, (ignoring some stray points) the curve has some significant value towards the understanding of ductile brittle transition depending upon grain size at constant temperature (R.T.). It clearly indicates that the materials having grain size numbers above 5 is ductile in nature while below 5 it is rather brittle. This result is also supplemented by the micrographs and photograph of cracked surface (Fig.5.6 and 5.7) showing two distinct types of mechanism working in the ductile and brittle range for the failure of the material. In mild steel (b.c.c. metal) due to yield drop large avalanches of dislocations are formed suddenly and cracks may be formed in a time too short to start slip from nearby dislocation source. This effect is more prominent in bigger grains as the area for dislocation pile up is much (see. 2.3.2.). Here the cracks formed along the grain boundaries at an angle of 45° to the tensile axis where shear stress is maximum. These cracks are responsible for the brittle type failure. In smaller grains the pile-up stress is not significant enough to initiate cracks.

There are some difficulty in viewing macrocracks because polishing powder might be trapped in their cavities. To get rid of this one should use ultrasonic bath which is not available. In fact scanning electron microscopy could have been very useful for this type of study.

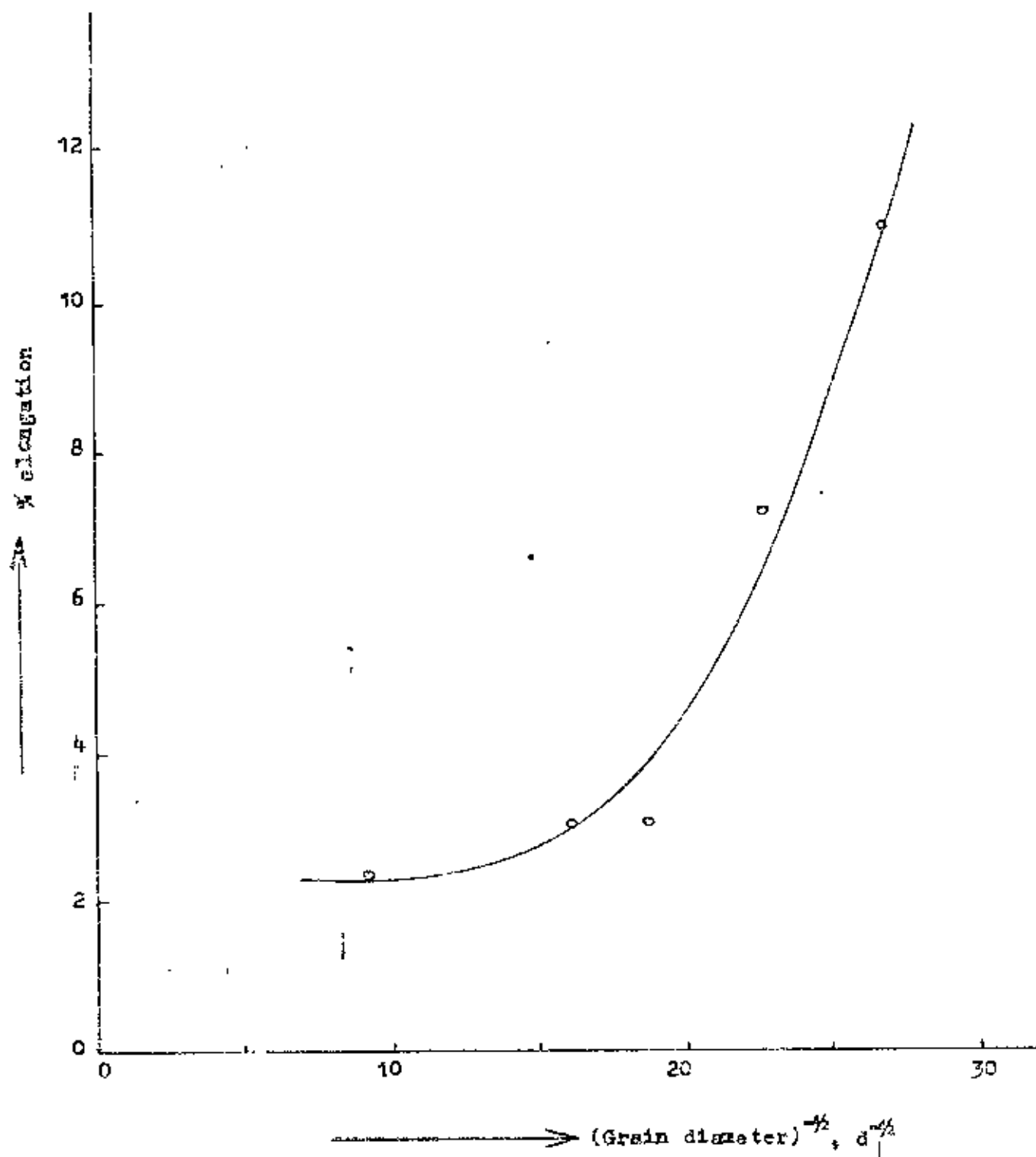
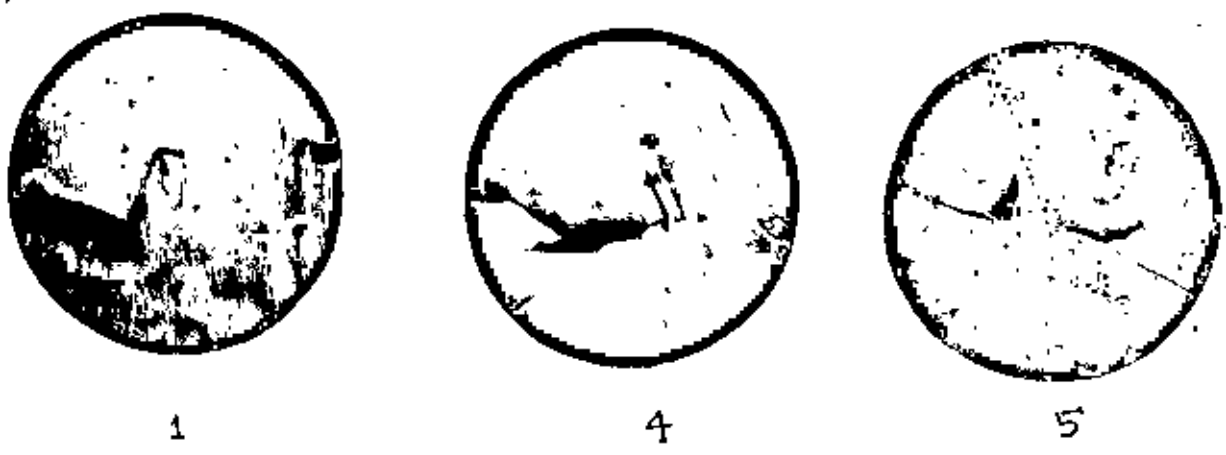
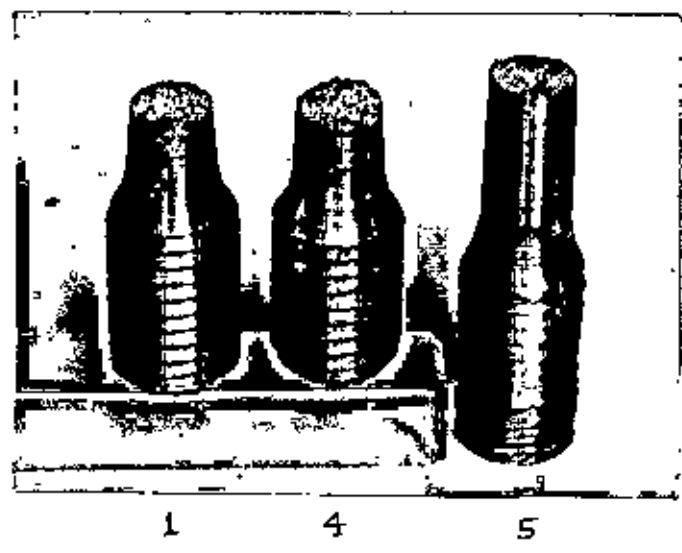


Fig.5.5 Effect of grain size on ductility.



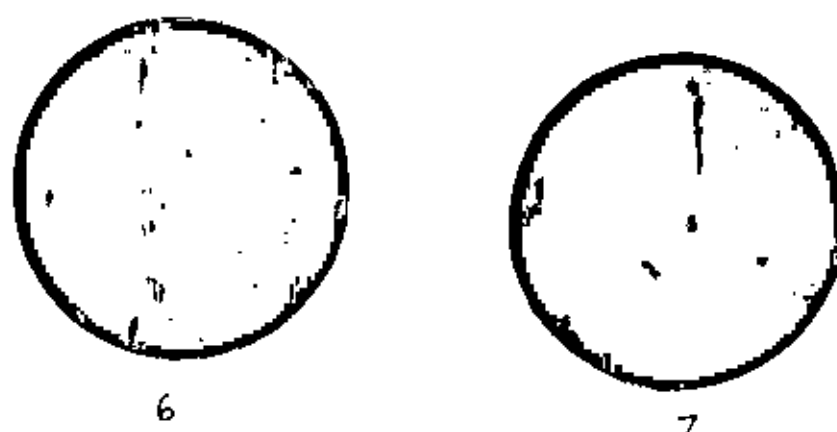
(a)



(b)

১৩৩
★
৬

Fig.5.6 (a) micrograph and (b) photograph of cracked surfaces of samples (grain size No. 1, 4, 5) showing brittle fracture.



(a)



(b)

Fig.5.7 (a) micrograph (b) photograph of crackled surfaces of samples (grain size 6 and 7) showing ductile fracture.

From the discussion the following conclusions may be drawn for 0.35% carbon steel.

1. Smaller grains (ASTM No. 6-8) and very big grains (ASTM No. 1,2) show less sensitivity towards growth with increasing temperature where as the rest (ASTM No. 3-5) grow significantly faster with the increase of temperature.
2. Since different grain sizes are obtained by varying the annealing temperature, the internal stress developed due to quenching also vary. The effect of internal stress is more pronounced on hardness than that caused by the variation of the grain sizes and bigger grains show higher hardness.
3. Yield strength increases with a $d^{-1/2}$ confirming Hall-Petch equation.
4. Tempering at 600°C shows no change in the trend of the variation of yield strength with grain size. This indicates that the influence of grain size on yield strength is more than that of internal stress or structural changes.
5. Yield strength increases at a faster rate similar to the increase in grain boundary area with decrease in grain size.
6. From the nature of fracture (necking, elongation, microcracks) it is found that there is a brittle-ductile transition depending on the grain size. Specimens having grain size numbers above 5 is ductile in nature while that of below 5 shows brittle characteristics.

Materials of different carbon percentages with identical tests will give a broader understanding of the properties of hypoeutectoid steels. Different grain sizes are obtained by varying the annealing temperature. These have an adverse effect on hardness. The real effect of grain size on hardness could be confirmed if the grain-sizes were obtained at constant temperature by changing the holding times.

There is a big scope for future work related to this field. The effect of grain size on creep and fatigue properties will definitely be very useful. Relation between strain rate and grain size may also be of value. For scientific interest a comparative study of the effect of grain size on b.c.c. , f.c.c. and h.c.p. metals may be made where the movements of dislocations will vary.

Appendix -1

From table 4.3 we can see that the sample which was annealed at a temperature of 1150°C prior to quenching, obtained maximum hardness which was 31.8 Rc. This quenched specimen was tempered at various temperatures and then their hardness were measured which is given in the table A1-1:

Table A1-1 specimen tempering treatment:

No.	Tempering Temperature $^{\circ}\text{C}$	Tempering time hr.	Cooling Method	Hardness Rc.
1	250	2½	Furnace Cooling	28.4
2	300	2½	" "	27.2
3	400	2½	" "	27.2
4	500	2½	" "	14
5	550	2½	" "	10.5
6	600	2½	" "	7.8

Table A1-1. Specimen tempering treatment :

The result of table A1-1 were represented graphically in Fig.A1-1 showing the relation between the tempering temperature and the hardness measured. The curve shows that hardness decreases with the increase of tempering temperature at constant time. So the fall of hardness with increasing holding time need not be considered.

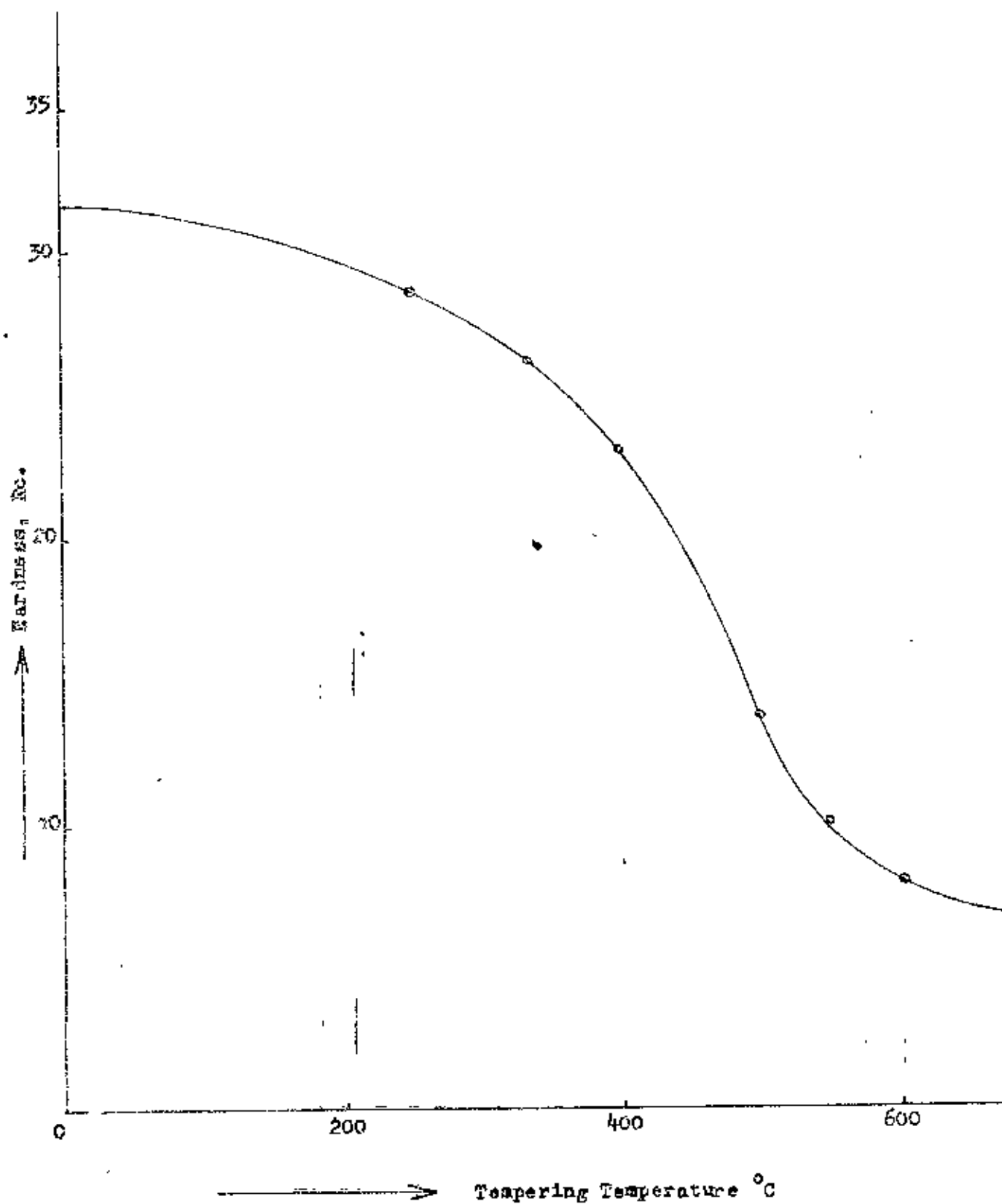


FIG-A1-1 Effect of tempering temperature on hardness.

Initially the decrease is rather slow with temperature where internal stresses due to quenching are relieved and diffusion processes for structural changes are not significant. Then there is a sharp fall in hardness due to structural changes. Later at a tempering temperature of 600°C the structure seems to be stabilized and further rise in tempering temperature shows little effect on hardness.

Appendix -2

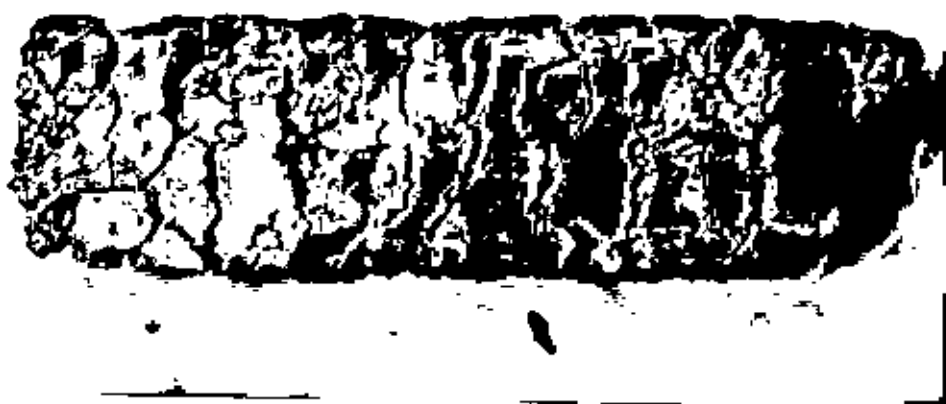


Fig-A2-1 Cracked billet.

The figure above shows a billet from Chittagong Steel Mills Ltd. which cracked unusually while it was being hot rolled in Bengal Rolling Mills, Tejgaon, Dacca. They said similar effect was also observed by other rolling mills using the same batch of materials. Of course besides grain size there may be other factors for the formation of these cracks.



References

1. E.O.HALL
Proc.Phys.Soc. London. 64B, (1951) P 747.
2. N.J.PETCH,
J.Iron St. Inst. 173, (1953)P 25.
3. A.CRACKNELL and N.J.PETCH
Acta Met 3(1955) P 186.
4. J.HESLOP and N.J.PETCH
Phil. Mag 1(1956) P 866
5. J.HESLOP and N.J.PETCH
Phil Mag 2 (1957) P 649
6. J.HESLOP and N.J.PETCH
Phil Mag 3(1958) P 1128.
7. N.J.PETCH
Conf.on the atomic mechanism of fracture,
Swainscott, (April 1959) P 54.
8. I.CODD and N.J.PETCH
Phil.Mag 5 (1960) P 30
9. A.A.JOHNSON
Phil Mag. 7(1962) P 177.
10. C.CRUSSARD
Conf. on Sixieme Colloque de Metallurgie, Centre d'Etudes
Nucleaires de Saclay (1962).
11. J.C.F.LI
Trans.Met.SOC: AIME 227 (1963) P 239.

12. R.W.ARMSTRONG and N.J.PETCH
Westinghouse Research Laboratories Scientific Paper
10-01000-0-P1, Dec 29 (1959)
13. O.E.JONES and J.R.HOLLAND
Acta Met. 16 (1968) P 1037.
14. H.H.TJENKSTRA
Acta Met 9 (1961) P 259.
15. B.JAOUL
J-Mech.Phys.Solids 9 (1961) P 69.
16. A.R.BALLY
A Text Book of Metallurgy
Macmillan, London (1967) P 1.4.
17. A.H.VAN BLACK
Elements of Material Science
Addision-Wiley, Massachusetts(1964) P 58-61, 95, 139.
18. R.E.REED HILL
Physical Metallurgy Principles
D.Van Nostrand, Princeton (1964) P 199-206.
19. C.S.SMITH
ASM SEMINAR
Metal Interfaces (1952) P 65.
20. J.E.BURKE and D.TURNBULL
Recrystallization and grain growth
Prog. in Metal Phys. 3(1952) P 220.
21. R.L.FULLMAN
ASM Seminar
Metal Interfaces (1952) P 179.

22. D.HARKER, and E.A.PARKER
Trans. ASM 34 (1945) P 156
23. P.FELTHAM and G.J. COPLEY
Acta Met. 6(1958) P- 539.
24. A.H.COTRELL
An Introduction to Metallurgy
Edward Arnold, London (1964) P 362.
25. A.H.COTRELL
The Mechanical Properties of Matter
John-Wiley, New York (1964) P- 362.
26. A.A.GRIFFITH
First International Conference.
Appl. Mech. Delft (1952) P 55A.
27. A.H.COTRELL
Trans. Met. Soc.
AIME 212 (1958) P- 192.
28. A.H.COTRELL
Fracture
John-Wiley, New York (1960) P-20.
29. G.L. KEHL
Metallographic Laboratory Practice
McGraw Hill, New-York (1949) P-270.
30. ASTM Standards
Part 1-A and 1-B (1946).
31. ASM. Metal Hand Book
Cleveland (1948) P. 401.
32. P.A. BECK
Jour Appl. Phys. 19 (1948) P. 507.

