

**MICROSTRUCTURES AND MECHANICAL
PROPERTIES OF WELDED COLD – ROLLED
LOW – CARBON STEEL PLATE**

BY

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MME/92/3677**

SUBMITTED TO

**THE DEPARTMENT OF METALLURGICAL AND
MATERIALS ENGINEERING**

**IN PARTIAL FULFILLMENT FOR THE AWARD OF
MASTER OF ENGINEERING IN THE DEPARTMENT
OF METALLURGICAL AND MATERIALS
ENGINEERING**

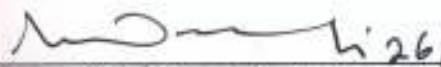
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MARCH, 2007.



CERTIFICATION

I certify that this research was carried out by **KUTELU BOLARINWA JOHNSON** at the Department of Metallurgical and Materials Engineering Federal University of Technology Akure, under our supervision and it has not been submitted elsewhere for the award of a degree

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DEDICATION TO:

- *My late Mother Mrs. Julianah Kutelu the fine sewing mistress*
- *My late Sister Mrs. Folasade Orimolade*
- *Those who are successful and want the success of others.*

ACKNOWLEDGEMENT

I wish to express my gratitude to Engr. (Dr.) Ben. O. Adewuyi, my project supervisor who has consistently lifted my spirit up in love, compassion and direction and whose constructive criticism, useful suggestion and encouragement had in no small measure resulted to the successful completion of this work.

I would like to acknowledge Dr. J.O.T Adewara, the head of Department (H.O.D) whose personality assisted me in going ahead with this work despite the fear.

I am also highly indebted to Engr. Omotoyinbo who despite his tight schedule spared his time in reading through the report; his suggestions and positive criticism were unparalleled.

I want to thank all other departmental lecturers, inclusive are Dr. J. Borode, Engr. Akinlabi Oyetunji and Mr. Alaneme K.K who have assisted me in one way or the other. At this junction, I must not fail to relay my unalloyed thanks to Mr. Olaniran Oladapo whose moral support was instrumental to the timely approval of the work for defence ; I thank Mr. Ademola Adelola and Mrs. Bunmi Iyamah for allowing me unlimited access to the use of their system, my thanks goes to Miss Faith Daniel who willingly and patiently typed several draft of the work.

My special appreciation goes to my Uncle Festus Olaiya Kutelu and my brother Oladapo Kutelu for their prayers, moral and financial assistance.

Lastly but no means the least, I would like express my sincere appreciation to Sola, my wife for her patience, counsel and contributions, physically, spiritually and otherwise and to my daughter peace Oyindamola who waited to be born some weeks after this investigations were complied.

Above all, honour and glory to the Almighty God, for his divine presence protection, provisions and his unparalleled contributions to the success of the work.

I am grateful to all.

ABSTRACT

Low-carbon steels have as one of their properties good weldability as a result they are often joined into useful structures by welding, however their safe use under different operating conditions is to a large extent determined by the microstructural characteristics of the welded parts, it is against this backdrop that investigation of the microstructures and mechanical properties of welded cold -rolled low- carbon steel plate was conducted, to achieve this low-carbon steel plates were machined and cut into various sizes. Fifteen (15) of these samples were normalized while another sets of fifteen (15) were annealed using laboratory heat treatment furnace. The normalized and annealed samples were heat treated at 920°C by cooling in air and furnace respectively to produce the desired normalized and annealed structures. The as-received samples were not heat treated so as to be used as control for the heat treated samples. The normalized, annealed and as-received structures were further cold worked by rolling at 5%, 10%, 20%, 30% and 40% percent reduction using laboratory rolling mill. Subsequently, the rolled samples were cut into two equal halves with hacksaw and welded together using electric arc welding process. From the welded structures, specimens for hardness, tensile strength and impact (notch toughness) energy tests and microstructural studies were produced by machining. Mechanical properties were determined at ambient temperature (25°C) and results analyzed graphically and microstructural characteristics obtained using optical microscope at magnification of 400xx. The microstructures revealed predominantly mixed phases of pearlite and ferrite, with the pearlite phase in ferrite matrix. Mechanical properties of the various structures were found to increase with increasing percent reduction with the non welded and welded parts of normalised structure exhibiting higher hardness and tensile at all ranges of percent reduction over that of corresponding parts of the annealed and as-received structures. Also, the welded part of normalised structure exhibited higher impact over that of the as-received structure at all ranges of percent reduction and higher impact over welded part of the annealed structure at lower percent reduction, but lower impact at higher percent reduction. The observed low impact (notch toughness) energy absorbed on impact by the welded part of the normalised structure at higher percent reduction is attributable to more fraction volume of pearlite phase in ferrite matrix.

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LIST OF SYMBOLS

YIELD STRESS	σ_0
FRACTIONAL STRESS	σ_1
EXTENT OF DEFORMATION	K_y
AVERAGE GROWN DIAMETER	D
TIME	t
CONSTANTS	k AND n
CHANGE IN STAIN	ϵ
ANGLE BOUNDARY	θ
ARC VOLTAGE	V
CURRENT	I
PROPORTION OF ARC ENERGY TRANSFERRED AD HEAT TO WORK PIECE	η
WELDING SPEED	v



LIST OF ABBREVIATIONS

HY	HIGH YIELD
AIS	AMERICAN INSTITUTE OF STEEL AND IRON
HSLA	HIGH – STRENGTH LOW – ALLOY
ASTM	AMERICAN SYSTEM OF TESTING AND MEASUREMENT
FCC	FACE – CENTER CUBIC
BCC	BODY – CENTER CUBIC
HAZ	HEAT – AFFECTED ZONE
BHN	BRINNEL HARDNESS NUMBER
ICRCG HAZ	INTERCRITICAL COARSE GRAIN HEAT – AFFECTED ZONE
LBZ	LOCAL BRITTLE ZONE
TP ²	SECOND – PEAK REHEATING TEMPERA TIRE
FS (A)	LAMELLAR STRUCTURES
HRC	ROCK WELL HARDNESS SCALE C
AWS	AMERICAN WELDING SOCIETY
DC	DIRECT CURRENT
AC	ATTENATING CURRENT
PCT	PERCENT
HRB	ROCK WELL HARDNESS SCALE B
YS	YIELD-STRENGTH
TS	TENSILE-STRENGTH
PF	POLYGONAL FERRITE
FG	FIND GRAIN
CG	COARSE GRAIN

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CHAPTER ONE



1.0 INTRODUCTION

The structural strength of a material is a complex of mechanical properties which can ensure reliable and long operation under specific operating condition, many of the properties of steels, especially those that are technologically important are determined by the *microstructure*, the use of a material (steel) by man for structural engineering application therefore, depends heavily on his ability to produce a particular microstructure with desirable properties in the steel when it has been fabricated into useful objects. There exists a close relationship between the structure and the properties of the steel, the structure is established by the past history of the steel.

Steels' structures allow prediction and evaluation of their properties and appropriate selection for specific application under particular force, temperature and environmental condition. The structure of the crystal lattice gives an indication of deformability characteristics of the steel, the grain size is particular important to the strength, ductility, hardness and toughness of the steel at room temperature.

Understanding the structure of steel together with a large element of empirical development, considerable improvement in steels' properties have been achieved chiefly by alloying and heat treatment (Kamma, 1993). The former involves the use of alloying elements, which are added to improve upon specific properties which are lacking in the metal, while the use of heat-treatment involves mainly the application of heat to bring about modification in the composition, arrangement and distribution of phases in the metal/alloy (Rejan et al. 1988). Recently Alaneme (2002), adopted a rather unorthodox method for alloy development by thermomechanical treatment and investigated the recrystallisation of low alloy steel in the $(\alpha + \gamma)$ phase region.

thermomechanical treatment and investigated the recrystallisation of low alloy steel in the $(\alpha + \gamma)$ phase region.

Frequently, the engineer is only concerned with the mechanical properties of the metals he uses with little or no consideration of some factors that have tremendous effects on its ultimate behaviour, some of which are due to forming and/or fabrication process among these is welding. The relative limited service performance of welded parts of steel structures as compared to non-welded parts under the same operating condition due to failure initiation in the welded area makes the effects of welding as critical as to inform considerable metallurgical research. Low carbon steels are generally very suited for light duty general engineering purposes such as welded structures, riveted structures and machined parts. Their wide use for structural applications in the oil and gas industries such as in the construction of oil ridge, oil platforms and pipelines are not only in their fair strength but also in their excellent formability and good weldability. However, their safe use under different loading conditions is determined by the effect of welding on their microstructural characteristics.

Studies conducted so far revealed that some related works have been carried out in this area of research by some researchers. Couque et al, (2001) investigated the fracture behaviour of a plain carbon steel with various microstructures at two loading rate corresponding to quasi-static and high rate loading (charpy impact test) and over a range of temperature. The effects of temperature, loading rate, and various microstructural parameters on the initiation of plane-strain fracture were explored. Zurek et al. (2004) studied the influence of dynamic spall loading on the cleavage-fracture behavior of a mild-carbon steel. Lee et al (2002) characterized fracture toughness under dynamic and quasi-loading over a wide range of temperature and investigated the effects of microstructure on the fracture toughness of an HY-100 steel. In a recent work,

Wang et al, (2004) studied effects of loading rates on fracture behaviour of low - alloy steel with different grain sizes., Sunghak et al (1993) carried out fracture toughness analysis of heat – affected zones (HAZs) in a high strength low-alloy steel welds. Q. xue et al (2003) investigated the constitutive response of welded HSLA 100 steel and studied the dynamic mechanical behaviour of a welded joint of high-strength, low alloy (HSLA) 100 steel by both quasi-static (103S^{-1}) and high strain rate (103S^{-1}) tension loadings at ambient and low temperatures.

. Despite the fact that cold deformation remains an important production route of crucial issue in many applications that require better surface finish, accurate dimension and an improved strength, investigation so far has not sufficiently revealed enough study on strength behaviour of welded steel at different degrees of cold deformation. Hence, the main concern in these studies is to carry out mechanical tests such as hardness, tensile, yield, impact and microstructural analysis on welded area of a low carbon steel plate at various percent reduction at ambient temperature (25°C), using base plate (non welded area) as control.

This work will not only provide framework for further research on this subject matter, but will also assist the metallurgical engineers generally in Africa and especially in Nigeria to understand more fundamentally the effect of welding at various degrees of cold deformation on the microstructures and mechanical properties of low-carbon steels. It is hoped that this understanding will aid in the production of local steel equivalent with optimal mechanical properties for indigenous oil and gas industries.

AIM

This work aims at providing framework for investigating the microstructures and mechanical properties of welded cold-rolled low-carbon steels used in oil and gas industries.

OBJECTIVE

The objectives are:

- I) produce annealed and normalised structures
- II) Carry out cold deformation at different ranges of percent reduction and subsequently welding operation
- III) Determine the microstructural and mechanical properties of the welded structures



CHAPTER TWO

2.0 LITERATURE REVIEW

The most important property from the point of view of mechanical engineering is the strength; this is strongly influenced by the microstructure and is thus one, but by no means the only microstructural sensitive property [Kalpakjian, 1995].

Haasan (1986) define microstructure from the thermodynamic point of view as all existing lattice defects in metal, which are not in thermodynamic equilibrium (according to type, number, distribution, size and shape).

Microstructure which as normally used covers structural features in the size range from atom (0.3nm) up to external shape of specimen at a size of millimeter (mm) or metres (m). The structural features include the composition, the crystal structure, the grain size, the size and distribution of additional phases [Martin, 1977].

A very high proportion of steels used in industry have a ferrite – pearlite structure. These include a wide range of mild steels [Lindberg, 1990].

2.1 MILD STEEL

Mild steels are steels with low carbon content and less than 1% total alloying additions. They include those in the American institute of steel and iron (AISI) series C – 1008 to C – 1025. Carbon ranges from 0.05 to less than 0.3% manganese ranges from 0.25 to 1.5%, phosphorous is 0.04% maximum, and sulphur is 0.05% maximum..

There are two important groups of mild steels – Dead mild steel and structural mild steel.

2.1.1 DEAD MILD STEEL

Dead mild steel has carbon content in the range 0.05% to 0.15%. It is produced as hot and cold worked sheet strip, rod, wire and tube is available in the hot worked or process annealed

condition. The steel has excellent ductility, which is essential for deep drawing of sheet, but the strength is low. The corrosion resistance of the steel is low but can be improved by plating with tin (tin plating) or with zinc galvanizing. Traditionally this steel has been used for car body panels, with a carbon content of $< 0.13\%$ to achieve a reasonable combination of yield strength ($> 140\text{Mpa}$) and formability ($n > 0.16$) other use of this steel is as packaging steel, for the manufacture of refrigerator, washing machines office furniture, pressings, reverts and nail. The manganese content of the steel is restricted to an amount (about $0.6\text{-}1.0\%$), which will form manganese sulphide (MnS) so as to prevent hot-shortness due to the presence of iron (II) sulphide (FeS) inclusion.

Pickering (1978), has pointed out that the non-metallic inclusion should be as low as possible, i.e. sulphur content should be low because non-metallic inclusion are detrimental to cold-formability.

2.1.2 STRUCTURAL MILD STEEL

Structural mild steel has carbon content in the range 0.10% to 0.25% . As the carbon is raised above 0.2% , the strength increases into the range required for constructional purposes, but ductility decreases (Davies et al, 1983).

The fabrication qualities (working and welding characteristics) of this steel are very good. The relative low strength and high ductility of these steels make it possible to cold-work them in the ferritic condition, when cold-rolled they give smoother finishes, closer tolerance, increased strength, improved fracture toughness and are easily welded with any of the arc, gas and resistance welding processes with electrode E60xx and E7xx as described in American Welding Society (AWS) specification. [www-key-to steel.com. 2005].

The steel is hot-rolled into plate for making boilers ships and vehicles and also into sections (e.g. girders, beams, joints) for use in bridges and buildings. The steel is often used in the hot-rolled condition but the smaller sized material may be normalized.

Although mild steel has excellent formability, their low strength requires relatively thick sections for a given load bearing application, new compositions of ferritic steel combined with new processing have significantly reductions in vehicle weight by permitting the use of thinner sections. One approach to improve low-carbon ferritic steels has led to the development of a class of steel referred to as high-strength low-alloy (HSLA). HSLA steels are alloyed with small amount of strong carbide-forming elements, such as vanadium and niobium, Mukerjee et al (2001) reported that very fine carbides are formed when the microalloyed steels are finish-hot worked low in the austenite phase field. The alloyed carbides limit the growth of the austenite. The resulting very fine ferrite grain size, together with fine alloy carbides, significantly increase strength.

Another approach to produce high strength and good formability in low-carbon steel is to heat a steel with a ferrite-pearlite microstructure into the ferrite plus-austenite phase field. Such intercritically annealed treatment is referred to as dual-phase steels intercritically annealed treatment converts the pearlite areas into pools of austenite and by controlling cooling rate or alloying the austenite or both transforms to structures other than pearlite on cooling [Honeycombe et al, 1998].

In the early 1950s, the work of Hall and Petch revolutionized the design of structural steel with the concept that refinement of ferrite grain led to an increase in both the yield strength and toughness of ferrite-pearlite steels. Thus, steel with yield strength and values up to about 300N/m^2 could be produced in aluminum-grained-refined compositions with good impact

properties and with good welding characteristics. Ferrite grain refinement remains the single most important metallurgical parameter in the make-up of modern structural grades.

Large grains in metals can be converted to fine grains by recrystallisation process taking place when a heavily cold worked material is annealed [Rollason, 1993].

2.2 RECRYSTALLISATION

When a strain-hardened metals is heated, the old grains are not restored, instead entirely new grains are formed, and their size may differ essentially from that of the initial grains. The formation of new equiaxed grains in place of the oriented fibrous structure is called *recrystallization* treatment or primary recrystallization. If any of the new grains have preferred condition for further growth, this stage of recrystallization is said to be secondary.

The greater the amount of prior deformation, the lower the temperature for the start of recrystallization. Since there will be a greater distortion and more internal energy left [Sidney 1982].

The size of the grain after cold working and recrystallization may be larger or smaller than that of the initial grain. The grain size depends upon the temperature of recrystallization annealing Fig. 2(a), the holding time at this temperature Fig. 2(b), the degree of previous deformation Fig. 2(c) the chemical composition of the alloy and the size of the initial grains.

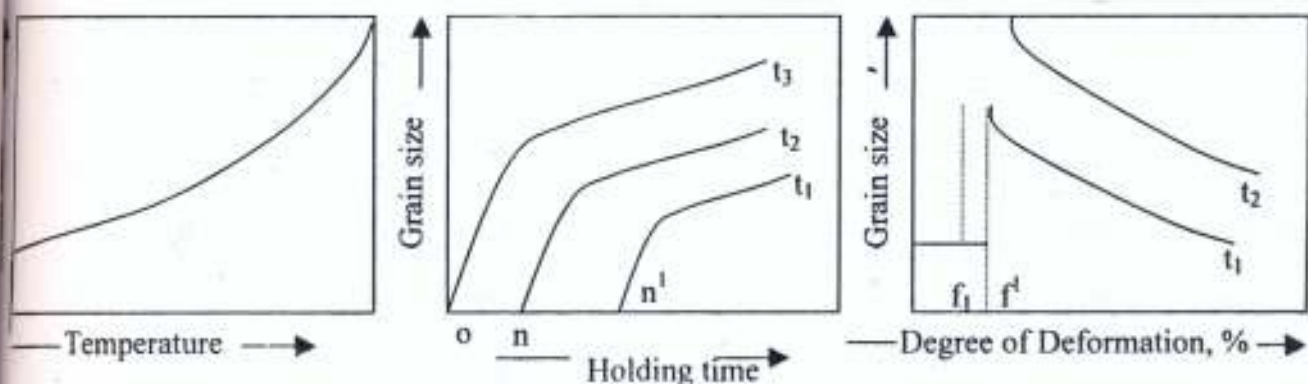


Fig 2 (a-c): Schematic representation of the effect of the temperature, holding time and degree of previous deformation on the size of the recrystallized grains

Fig 2 (a-c): Schematic representation of the effect of the temperature, holding time and degree of previous deformation on the size of the recrystallized grains (θ , kt_2 , kt_3) and on lengths of the incubation periods f and f^1 (critical degrees of deformation)

For a given degree of deformation, the grain size increases with a rise in temperature and an increase in holding time. The greater the degree of previous deformation, the smaller the recrystallized grains. At low temperatures (t_1 and t_2) in Fig. 2(b), but above recrystallization temperature, t_n , the formation of recrystallization grains does not begin at once, but only after a certain time interval called the incubation period.

At very low degrees of deformation Fig. 2(c), heating does not cause recrystallization. According to Lakhtin (1979), at a deformation of 3 to 15 percent, the grain size grows rapidly following annealing and may become many time larger than that of the initial grains. Such a degree of deformation is said to be critical. When a metal is subjected to the critical degree of deformation recrystallization does not proceed by the mechanism in which new grains are formed and their subsequent growth. Heating following the critical degree of deformation leads only to a rapid growth of some of the initial unrecrystallization grains by absorbing adjacent ones. This mechanism of recrystallization is due to the non-uniform deformation of various grains at low degrees of deformation. Upon deformation at values above the critical degree the metal undergoes primary recrystallization. Hence, the critical degree of deformation is the minimum one, above which primary recrystallization is possible in subsequent heating of the metal.

This derives from the fact that the number of regions with higher dislocation density increases with the degree of deformation, thereby increasing the formation of centers of

recrystallization. At high degrees of deformation, the rate of formation of recrystallized nuclei exceeds their rate of growth. This enables a fine-grained structure to be obtained.

The dependence of yield stress on grain size is expressed by the well known-Hall-Petch equation.

$$\sigma_y = \sigma_i + K_y D^{-1/2} \dots\dots\dots 2.0$$

Where σ_y is the yield stress, σ_i frictional stress opposing motion of a dislocation, K_y the extent to which dislocations are piled up at barriers, and D is the average grain diameter. Therefore, the finer the grain size, the higher will be yield stress. In fact, grain refinement is the only commercially available conventional technique, which improves strength and ductility at the same time [Rajan et al, 1988].

Grain size significantly influences the mechanical properties of metals. At room temperature, a *large grain* is generally associated with low *strength, hardness, and ductility*. Large grains, particularly in sheet metals, also result in a rough surface appearance after being stretched.

Fujioka et al, (1979) investigated the effect of grain size on the creep behaviour of steel and reported that at low temperature, the rupture life increases with decrease in grain size whereas at high temperature, it decreases with decreasing grain size.

Grain size has marked influence on the impact transition temperature. An increase in grain size raises the impact transition temperature, and this makes the steel more prone to failure by brittle fracture. The effect of grain size on the impact transition temperature is more pronounced in the case of low carbon steels.

Coarse grains steel have better hardenability than fine grained steel. This is because coarse-grained steel has fewer grain boundaries. Since grain boundaries are the region of high

rate of diffusion. Consequently, formation of pearlite, which is a diffusion controlled process, starts at grain boundaries with smaller grain boundary area in coarse grained steels, preferential formation of martensite from austenite takes place on cooling. It will further be relevant to note that pearlite formation proceeds by nucleation and growth mechanism. Austenite grain boundaries act as nucleation site for pearlite formation.

Coarse grained steels have better machinability than fine grained steels. Coarse grained steel has reduce toughness, which is understood to be the source of providing small discontinues chip during machining [Sidney, 1982]

Grain boundaries have an important influence on the strength and ductility of metals. Because they interfere with the movement of dislocation, grain boundaries also influence strain hardening. These effect depend on the temperature, rate of deformation and the type and amount of impurities present along the grain boundaries. Grain boundaries are more reactive than the grain themselves. This is because the atom along the grain boundaries are packed less efficiency and are more disorder and, hence have a higher energy than the atom in the orderly arrangement within the grains [Kalpakjian 1995].

Commercial metal are never " pure" , and due to minor segregation at the ingot casting stage, a proportion of the impurities are deposited are along the grain boundaries [black et al ,1996].

2.3 GRAIN GROWTH

Grain size depends both on time and temperature. Other factors, however, are known to influence the migration of grain boundaries. These included the concentration of solute element at grain boundaries which can introduce a frictional drag on the moving boundary and second phase particles, which decrease the grain-boundaries area, hence the overall grain-boundaries

energy (Cahn,1962). Miller (1951), studied the kinetics of the grain growth in the austenite range and found that at any given temperature the average grain diameter (D), depend upon the time (t), according to the equation:

$$D= Kt^n.....(2.1)$$

Where K and n are constants.

Grain growth of austenite can be modified considerably by the addition of element such as aluminum, niobium, vanadium, and titanium, e.t.c all of which produce small, sparingly soluble precipitates.

There are two principal type of grain growth coarsening curve as pointed out by Bain and Paxton (1947). Plain carbon steel without for example, aluminum addition, show gradual growth of all grain as the temperature is increased. The other type of grain growth is that observed in steel containing addition of grain refining elements. In this case the grain size remaining small (ASTM6-8) at temperature up to the grain coarsening temperature only a few of the grain growth resulting in mixture of coarse and fine grains. When the temperature is rising sufficiently, all the grain become coarse (ASTM0-2) this type of grain growth may sometimes be referred to as secondary recrystallization.

The time exponent n, in equation (2.0), has a value of 0.25-0.4 (Nielsen, 1996). Thus very rapid growth is followed by a period where the grain size curves become almost asymptotic to a certain (temperature dependent) value. It is difficult to control the length of a recrystallization process to give a grain size to which the grain will grow by other means.

Rashid, (1975) investigated that, a mixture of ferrite and martensite could be produced in a 0.15% carbon steels annealing in the intercritical, two phase ferrite plus austenite region, between AC1and AC3, by holding in two phase carbon can diffuse from ferrite to a more

accommodating austenite region, rising the carbon content of the austenite to a level significant higher than that of the normal base composition. In turn this increase the hardenability of the austenite allowing martensite to form on cooling to ambient temperature.

Important factors like composition, heat-treatment and forming processes have tremendous effect on the ultimate behaviors of steel (Dollason, 1986).

2.4 PLASTIC DEFORMATION

In crystal, deformation can occur by slip or twining. Slip occur in a crystal lattice along plane and direction with the most closely packed atomic arrangement, where the shear strength σ is the lowest, and the stress γ is considerable, this is due to the distance between the adjacent atomic plane is maximal (i.e the bond are weakest between such planes)

Plastic deformation or cold working affects all properties of metal that are dependent on lattice structure. Tensile strength, yield strength, and hardness are increase. While ductility, as represent by percentage elongation, is decreased [Kalpakjian, 1995].

When crystal are deforms, there is some distortion of the lattice structure. This distortion is greater on the slip planes, and grain boundaries and increase with increase deformation. Much of the strength of metal is due to the fact that the slip plane lie in different direction in the different crystal boundaries atom are shape (i.e. have a position in the lattice of two or more crystal [Black et al, 1996].

One of the remarkable feature of plastics deformation is that the stress requires to initiate slip is lower than that requires to continue deformation on subsequent plane. Aside from distortion of the lattice structure, the pile-up of dislocation against obstacles (such as grain

boundaries and foreign atoms) and the locking of dislocation on intersecting slip plane increase the resistance to further deformation [Sidney 1982].

Lakhtin, (1979), investigation plastic deformation in crystal, his report was that in metal having a face- center cubic (FCC) such as austenite iron (γ -Fe) slip occur along the octahedral plane (111) and the direction of the diagonal of the face of the cubic [110], while in a centre body cubic (BCC) mostly readily accomplished along plane (110) and direction [111] of the body diagonal of the cube.

Bodin et al (2002) reported in their work on texture development in austenite during hot rolling and direct annealing that the deformation of austenite FCC metal occurs by twinning, their findings showed that the preferred slip system is always the same as the close-packed plane (111) slip in the close-packed direction [110] and twinning occurs at the (111) [112] system.

Lakhtin, (1979) studied the changes in the structure of a metal in a plastic deformation and reported via a microscope on polished deformed specimen that the first to be observed were traces of slip in the form of straight lines, similarly oriented within the limit of the separate grains.

Owing to slip processes, the grains (crystallites) were changed in shape when subjected to considerable deformation. The grains, prior to deformation were more or less equiaxed shape. As a result of displacements along slip planes, the grains were elongated by deformation in the direction of the acting forces γ , forming a fibrous or banded structure simultaneously with the change in the shape of the grains the slip blocks were broken up inside the grain and the angle of disorientation between them increases.

The changes in crystallographic orientation during deformation of grains are not random. They are a consequence of deformation occurring on the most favourably oriented slip or

twinning systems. As a result the deformed material acquires a preferred orientation or texture. If the metal subsequently recrystallises nucleation may occur preferentially in regions of a particular orientation. The ability of the nucleus to grow is determined primarily by the orientations of adjacent regions in the microstructure. Together, these features (nucleation and growth) ensure that a texture also develops in the recrystallized material. Such a texture is called a recrystallisation texture to distinguish it from the deformation texture from which it develops. In ferrite, with its BCC structure, no close-packed planes exist. The close-packed direction is the $[111]$ direction and the plane with the highest density is the (110) plane. Apart from the $(110)[111]$ slip system, the $(112)[111]$ and the $((123)[111])$ slip systems can also be activated [Bodin et al, 2002].

Bodin, (2002) reported that unlike the fully austenitic rolling, the deformation of the austenite in the two-phase structure will not result in new texture components, neither during recrystallisation of the austenite nor during transformation into ferrite. The deformation and subsequent cooling of an austenite – ferrite two-phase structure leads to crystal rotation by further deformation of the remaining austenite phase; (possibly) recrystallisation of the austenitic, crystal rotation by deformation of the ferrite phase to more stable texture component; (possibly) recrystallisation of the ferrite; and completion of the austenite-to-ferrite transformation.

2.4.1 HOT DEFORMATION

Deformation at these temperatures causes hardening “hot strain hardening”, which is completely or partially eliminated by *polygonization* and *recrystallisation* occurring at the working temperature and in subsequent cooling. In hot working of metals strengthening due to strain hardening (increasing the dislocation density) in the deformation process continually

alternates with the softening process (reducing the dislocation density) in dynamic polygonization and recrystallisation during deformation and subsequent cooling [Lakhtin, 1979].

When metal is deformed at a high temperature (hot deformation) the deformed microstructure usually recovers or recrystallises.

It is one of the functions of hot working to break down the large crystals which form due to slow cooling of the ingot, and to break up and disperse segregation, one result of the dispersion of segregations is that commercial forms of metal tend to have directional properties. Sheet and strip metals used in presswork will bend across the direction of rolling more successfully than along the direction of rolling; sharp bends made along the direction of rolling may cause cracks to appear on the tension side of the bend. [Black et al, 1996].

2.4.2 COLD DEFORMATION

A material is considered to be cold-worked if its grains are distorted after plastic deformation is completed [Sidney, 1982].

Cold deformation is carried out at temperature below that of recrystallisation temperature (i.e. at room temperature) therefore, cold deformation (working) leads to strain or work hardening of the metal [Lakhtin, 1979].

Low carbon content and minimum residual alloying content are required for cold formability

Low carbon steels in thin section sizes may be rolled at ambient temperatures. This breaks up the ferrite pearlite structures and causes work hardening recrystallisation cannot occur at these temperatures [Robert Ross, 1992].

If a metal or alloy is still in the strain-hardened state or has a polygonised structure following deformation at temperatures above recrystallization temperature, the process is considered cold working [Lakhtin, 1979].

It generally known that during cold deformation, near the boundaries grains are deformed to a lesser degree than at some distance away. Therefore, a cold deformation is essentially a shear internal to the grain (i.e. happening within the grain). The grain boundaries remain intact and mass continuity is maintained, the deformed metal exhibits greater strength because of the entanglement of dislocation with grain boundaries.

In cold-working a metal, the temperature is not higher than 0.3 of the melting point ($\leq 0.3 T_m$) of the metal. In some cases, however, the working may be done at temperature that will produce increased ductility and reduced strength, but will be below the recrystallisation temperature. Cold-worked structure forms high-dislocation density regions or *tangles*, which soon develop into tangled network. Thus, the characteristic structure of the cold-worked state is a *cellular substructure* in which high-density-dislocation tangles form the cell walls [Dieter, 1988].

Changes of mechanical properties which arise during cold working are mainly the result of deformation of crystals by "slip" so that ultimately the number of 'slip planes' remaining unused is small [Black et al, 1996].

Tensile strength, yield strength, and hardness are increased, while ductility, as represented by percent elongation, is decreased. Although both strength and hardness increase, the rate of change is not the same. Hardness generally increases, most rapidly in the first 10 percent reductions, whereas the tensile strength increases more or less linearly. The yield strength

increases more rapidly than the tensile strength, so that, as the amount of plastic deformation is increased, the gap between the yield and tensile strength decreases.

Recrystallized grain-size decreases with increasing cold reduction because of an increasing density of recrystallisation nuclei with larger rolling strains (Leslie, 1981).

Since in most cold-working processes one or two dimensions of the metal are reduced at the expense of an increase in the other dimensions cold-work produces elongation of the grains in the principal direction of working. Severe deformation produces a reorientation of the grains into a preferred orientation [Dieter, 1988].

For equal amount of cold working, more strain hardening is introduced into initially fine-grained than into initially coarse-grained metal. Therefore, the finer the initial grain size the lower the recrystallisation temperature. By the same reasoning, the lower the temperature of cold working, the greater the amount of strain introduced, the lower the temperature of cold working, the greater the amount of strain introduced, effectively decreasing the recrystallisation temperature for a given annealing time [Sidney, 1982].

2.5 STRAIN HARDENING

Hardening of the metal in the process of plastic deformation (strain hardening) is due to increase in the number of defects of crystal structure (dislocations, vacancies and interstitial atoms). All the defects impede motion of the dislocations, thereby increasing the resistance to deformation and reducing ductility [Lakhtin, 1979].

Kandakar et al, (1979) asserts that with an increase in the degree of deformation properties increasing the resistance of metal to deformation) are raised and, on the other hand, the capacity for plastic deformation (percent elongation δ and reduction in area Ψ), is reduced.

The amount of strain hardening introduced by a given amount of elongation increases as the grain size decreases [Sidney, 1982].

The increased in strength depends on the degree of deformation (percent reduction) to which the metal is subjected, the greater the deformation, the stronger the metal becomes. The increase in strength is greater for metal with smaller grains because they have a larger grain – boundary surface area per unit volume of metal [Lakhtin, 1979]. As deformation proceeds the metal becomes progressively harder and stronger and whether by slip or by twinning, a point is reached when no more deformation can be produced. Any further increase in the applied force will lead only to fracture. In this condition, when the tensile strength and hardness have reached a maximum, the material is said to be *work hardened*. As deformation proceeds the capacity for further deformation decreases and the force necessary to produce it must increase. All dislocation present at the start of stress application have moved into jammed positions, as have all new dislocation generated by the progressive increase in stress, until no further slip (maximum strength and hardness) further increase in stress would give rise to fracture. [Higgins, 1993].

The increase in strength is based on the concept that grain boundaries act as barriers to dislocation motion, which is brought about when a material is subjected to stress. Thus, smaller grain size i.e. increased in the number of grain boundaries leads to higher strength [Irvine, et al, 1962]

For equal amounts of cold-working, more strain hardening is introduced into initially fine-grained than into initially coarse – grained metal. Therefore, the finer the initial grain size the lower the recrystallisation temperature. By the same reasoning, the lower the temperature of

cold working, the greater the amount of grain introduced, effectively decreasing the recrystallisation temperature for a given annealing time [Sidney, 1982].

Reis et al, (2000) investigated the effect of an increasing strain on the substructural features of low-carbon steel by comparing the 95% cold rolled material with a sample that was cold rolled to a reduction of 70%. It was reported that, only the presence of the deformed bands can be observed on the microstructure of the latter and that these bands contain much less high angle boundaries ($\theta > 20^\circ$) than the ones obtained after 95% rolling strain. Hence, he concluded that a change of strain from $\xi = 1.2$ (70% rolling reduction) to $\xi = 3.0$ (95 rolling reduction) has indeed produced a significant increase of high-angle-grain-boundaries in the structure.

The as-cast crystal during cold rolling are disturbed as mechanical slip takes place and become work hardened (overall change in properties).

Dieta, (1988) reports that the greater is the force required for further deformation, in other words the specimen work hardens when the rate of hardening becomes zero, the specimen fractures.

The deformation behaviour of steels is affected by the grain size. Kandekar et al, (1979) investigated the deformation behaviour of an ultrafine-grained medium carbon steel. It was reported that yield drop increases with grain refinement, the total elongation remains unaffected, Luder's elongation and post-neck elongation increases, uniform elongation is reduced and the strain hardening exponent, n , reaches a near limiting value as the grain is reduced.

2.6 FRACTION VOLUME OF PEARLITE AND ITS EFFECT ON THE STRENGTH PROPERTIES OF LOW-CARBON STEELS

Pearlite occupies substantial smaller volumes of the microstructure of low-carbon steels. Honeycombe, (1998) reported that in these steels the yield strength is not markedly affected as

the proportion of pearlite is increased, provided other factors, e.g. ferrite grain size is kept constant. However, the tensile strength is quite sensitive to the pearlite content which is explained by the fact that there is a linear relationship between work hardening and the pearlite content which arises because pearlite work hardens much more than ferrite.

Robert Ross, (1992) asserts that the mechanical properties of annealed, normalized or hot-rolled steels are directly dependent on the amount of pearlite present in the microstructure.

Sidney, (1982) investigated the effect of pearlite on the impact transition temperature of low-carbon steel, he reported that the impact transition temperature is raised substantially as the pearlite content is increased.

The presence of pearlite in the microstructure provides sites of easy nucleation of cracks, particularly at the ferrite-cementite interfaces. However, as the crack can only propagate in ferrite, a short distance before encountering another cementite lamella energy is absorbed during propagation. The result is that there is a wide range of transition temperature. In contrast, the low energy absorbed overall in impact tests on pearlitic structures arises from the fact that many crack nuclei can occur at the pearlite interfaces which, together with high work hardening rate, restricts plastic deformation in the vicinity of the crack [Nat, 1992].

2.7 MICROSTRUCTURE OF COLD-ROLLED STEEL SHEETS

The response of microstructures to deformation is heavily influenced by the alloy compositions of steels.

Sung-Joon et al, (2001) investigated the microstructures of three kinds of cold-rolled steel ingots, ECO-1, ECO-2, and ECO-3. The three ingots were fabricated by vacuum induction melting and aluminium killing. The carbon content was fixed at 0.015wt pct and 0.15 to 0.25wt pct, respectively, and reported through optical micrographs of the cold-rolled ECO-1, ECO-2,

and ECO-3 steels that all the three steels have a mixed structure of ferrite and pearlite elongated along the rolling direction, showing a band structure, particularly as the manganese content increases or as the silicon content decreases, the formation of the band structure is accelerated; thus, the ECO-3 steel, with high manganese content is completely composed of band structure. It is generally known that the development of the band structure becomes stronger with high manganese contents. This is because the area where manganese and carbon are simultaneously segregated during casting are not completely homogenized, due to the reduced diffusion rate of carbon by the presence of manganese as well as the slow diffusion of manganese during homogenization; thus, they are elongated in a band-type manner wherein pearlite grain are intensively formed. Since this band structure works as a factor to deteriorate ductility, it can adversely affect the elongation enhancement when pearlite grains are completely decomposed during intercritical annealing.

Chung et al, (2001) reported from *in-situ* transmission electron microscope tensile experiments that retained austenite associated with nearby ferrite and bainite is transformed to martensite during plastic deformation, thereby working favourably for the ductility improvement. However, film-shaped retained austenite grains located at bainite or martensite lath boundaries and those isolated in ferrite grains are not transformed to martensite even when a considerable amount of plastic deformation is applied, thus not contributing to the ductility improvement. Chung interpreted the latter case by the size stability effect due to the extremely small size of retained austenite. The former case is associated with the deformation constraint effect as retained austenite are surrounded by bainite which are not deformed easily

Certain steels may develop maximum potential hardness even though they are cooled at a rate equal to, or greater than, the critical rate. This situation may be attributed to the presence of retained austenite in the quenched steel [Donald, 1987].

2.8 HEAT TREATMENT

The usefulness of steel is due largely to the relative ease with which its properties may be altered by controlling the manner in which it is heated and cooled. The changes that occur in the properties of the steel are directly related to changes in the structural make-up of the steel i.e. the nature and appearance of these transformation products determine the physical and mechanical properties of the given steel [Sidney, 1982].

The desired levels of mechanical properties can be obtained by altering the size, shape and distribution of various constituents. This is achieved in practice by the process of heat treatment [Rajan et al, 1988].

Heat treatment in general sense may be defined as an operation or combination of operations involving the heating and cooling of metal or alloy in the solid state for the purpose of obtaining certain desirable conditions or properties [Donald, 1987].

The principal kinds of heat-treatment used in practice which differently affect the structure and properties, and which are assigned to meet the requirements made to the semi fabricated materials (castings, forgings, rolled stocks) are : (a) Annealing (b) normalizing (c) hardening and (d) tempering .In all of these processes the steel is heated fairly slowly to some predetermined temperature, and then cooled. It is the rate of cooling which determines the resultant structure of the steel and hence, the mechanical properties associated with it. The final structure will be independent of the rate of heating, provide it has been slow enough for the steel to reach structural equilibrium at its maximum temperature. The subsequent rate of cooling,

which determines the nature of the final structure, may vary between a drastic water-quenched and slow cooling in the furnace [Lakhtin, 1979].

Time and temperature play important role in the heat-treatment especially in austenizing processes [Higgins, 1993].

Phase transformation or phase change may involve a change in the state of aggregation in the crystal structure or in composition [Davies, 1984].

The refinement of the grain size of ferrite provide one of the most important strengthening route in heat treatment of steel [Karlsson,1973].

An austenite grain transform to several ferrite grain on cooling where as a single ferrite grain transform to one austenite grain on rapid heating. Rapid austenizing minimize growth of austenite grains. When cooled below AC3 temperature the austenite in the hypoeutectoid range become supersaturated in carbon [Nicholson, 1970].

Ferrite normally nucleate at the austenite grain boundaries . on further cooling below A1 the carbon enrich region of remaining austenite provide nucleation site for pearlite. The ferrite grow from the grain of boundaries often appear in a plate -like shape (widmanstatten plate) in large grains, ferrite can nucleate also within the austenite grain and can grow as widmanstatten plate or blickly ferrite. Increase under cooling will enhance homogenous nucleate of ferrite relative to grain boundary nucleation producing further refinement of the ferrite- pearlite structure [Donald et al, 1969]

. Honeycombe et al, (1995) showed that most plain carbon steel are not subject to heat treatment in the sense of quenching followed by tempering, but they are cooled at different rate to obtain a range of structures two important heat treatment which have special but very precise meaning when applied to these steels are normalizing and annealing

2.3.1 NORMALIZING

Normalizing leads to complete recrystallization and eliminate the coarse grained structure obtain in casting or rolling or open-or-close-die forging. All unequal stress are removed although the result is seldom complete strain free for there will be some air hardening except in low mild steels, (www.key-to-steel.com2005).

The more cooling in air used in normalizing, cause the austenite to decompose at lower temperature. This increase the disparity of the ferrite – cementite structure and the amount of pearlite or, more exactly , the quasi- eutectoid (of the sorbite or troilite type). This raises the strength and hardness of normalized medium- and or high-carbon steel by 10 to 15 percent as compared to annealed steel [Lakhtin, 1979].

Normalizing differs from full annealing in that the metal is allowed to cool in still air. The structure and properties produced, however, vary with the thickness of metal treated. The tensile strength, yield point, reduction of area and impact value is higher than obtain by annealing [www.key-to-steel.com 2005].

The increase in cooling rate due to air cooling as compare with furnace cooling affect transformation of austenite microstructure in several ways [Sidney 1982]. The degree of hardness obtained during normalizing is proportional to material carbon content and alloying element as alloying has the effect of slowing the speed with which the iron carbide can return to equilibrium. [Robert Ross1992].

Normalizing alone or normalizing with high-temperature tempering is used for casting of medium-carbon steel instead of hardening and high temperature tempering. The resulting mechanical properties are lowered, but the work is warped less than in hardening and possibility of cracking formation is practically excluded.

Normalizing is used for high- carbon (hypereutectoid) steel to eliminate the cementite network that may develop upon slow cooling in temperature range A_{cm} to point A_1 . Normalizing followed by high temperature tempering at (600 to 650^o C) is often applied instead of full annealing to correct the structure of alloy steel because the productivity of two operation is greater than that of a single annealing operation [Raymond,1985]

2.8.2 ANNEALING

Annealing at fairly high temperature followed by cooling e.g. in furnace results in transformation which provides good machinability [Sidney, 1982].

During annealing ferrite structure cannot dissolve carbides, whereas these are taken into solution in austenite and at the same time there is a variation in volume in the change from ferrite to austenite. This 'recrystallisation' as it is called, completely eliminates all trace of previous thermal treatment, giving a fresh start. The slow cool from the critical temperature maintains equilibrium conditions at all times, allowing the iron carbide to return to its natural position, resulting in completely stress-free structure [Robert Ross, 1992].

Prolong annealing induces greater ductility at the expense of strength, owing to the tendency of the cementite in the strained pearlite to "ball-up" or spheroidize. This is known as "divorced pearlite". The ferrite grains also become larger, particularly if the metal has been cold worked to critical amount. A serious embrittlement sometimes results after prolonged treatment owing to the formation of cementite film at the ferrite boundaries with severe forming operations, cracks are liable to start at these cementite membranes.

The table below shows the average annealing and hardening temperature

Carbon %	0.1	0.2	0.3	0.5	0.7	0.9 to 1.3
Average temp. °C	910	860	830	810	770	760

Table 2.0: Average annealing and hardening temperatures [source: www.key-to-steel.com, 2005]

The temperatures allow for the effects of slight variation in the impurities present and also the thermal lag associated with the critical changes. After soaking at the temperature for a time dependent on the thickness of the article.

Consider the heating of 0.3% carbon steel. At the lower point (AC1), each "grain" of pearlite changes to several minute austenite crystals and as temperature is raised the excess ferrite is dissolved, finally disappearing at the upper critical point (AC3), still with the production of fine austenite crystals. This is necessary for the carbon to become uniformly distributed in this austenite. The properties obtained subsequently depend on the coarseness of the pearlite and ferrite and their relative distribution. These depend on:

- (i) The size of the austenite grain, the smaller their size the better the distribution of the ferrite and pearlite.
- (ii) The rate of cooling through the critical range, which affects both the ferrite and the pearlite.

As the temperature is raised above AC3 the crystals increase in size. On a certain temperature the growth, which is rapid at first diminishes. Treatment just at the upper critical point should be aimed at, since the austenite crystals are then small.

By cooling slowly through the critical range, ferrite commences to deposit on few nuclei at the austenite boundaries. Large rounded crystals are formed evenly distributed among the relatively coarse pearlite. With a higher rate of cooling, many ferrite crystals are formed at the

relatively coarse pearlite. With a higher rate of cooling, many ferrite crystals are formed at the austenite boundaries and a network structure of small ferrite crystals is produced with fine pearlite in the center. [www.key-to-steel.com2005].

When the steel is heated well above the upper critical temperatures large austenite form. Slow cooling gives rise to the Widemanstatten type of structure, with its characteristics lack of both ductility and resistant to shock. This is known as an “overheated structure” and it can be refined by reheating steel to just above the upper critical point. Surface decarburisation usually occur during the overheating [www.key-to-steel.com2005].

2.9 WELDING

Gourd, (1986) defines welding as the joining of two polycrystalline workpiece – usually of metals by bringing their fitted surfaces into such intimate contact that crystal-to-crystal bonding occur

A weld can be defined as a union between piece of metal at faces render plastic or liquid by heat or pressure or both. This union is the continuous load bearing metallic bridge between the piece of metal. Industrial welding usually entails local heat from a burning gas or an electric arc, or heat generated by electrical resistance [Agarwal, 1992].

The weld metal, the heat-affected zone, and eth thermal stresses and strain caused by welding may combine to yield a product different in strength, ductility, leak tightness and dimensional stability from that of the non-welded metal.

2.9.1 ARC WELDING

• In arc welding the first run of weld metal flowing on to the cool plate is virtually dull cast. The metal on the top of the weld area freezes quickly as the heat is removed from it and small chill crystals are formed. Below, these crystals grow away from the sides towards the

hotter region of the weld metal and thus growth is faster than the tendency to form new dendrites so that columnar crystals are formed. Because of the high temperature of the molten metal in arc welding, these crystals have enough time to grow. On each side of the weld is a heat-affected zone in which the temperature have been raised above the recrystallisation temperature and in which therefore grain growth has occurred. Beyond this zone, the base plate structure is unaffected [Davies, 1984].

2.9.2 GAS WELDING

In gas welding, the heat-affected zone will be wider than in arc welding because, although the flame temperature is below that of the arc, the arc is more localized and the temperature is raised more quickly. When the flame is applied, to the plate its temperature is raised so that chill casting does not occur. Grain growth will be more pronounced because the heat is applied for a longer period than in arc welding [Houldcroft, 1988]

2.9.3 HEAT-AFFECTED ZONE

Work-hardened metals are characterized by increase hardness and tensile strength resulting from deformation or working at room temperature. The effect of cold working can be removed by heating to a critical temperature above which softening occurs. This is known as *recrystallisation* [Gourd 1986]. During welding, the base metal experiences a range of temperature invariably from fusion to melting, it experiences a cooling rate of the same order of magnitude as quenching, softening and recrystallisation occur in the heat-affected zone (HAZ). The heat-affected zone (HAZ), which experiences temperature above this critical temperature softened, and its tensile strength falls below that of the unaffected parent metal, which still has the work hardened properties. It is not practicable to recover hardness in the heat-affected zone

The heat-affected zone (HAZ), which is cooled at different rates and includes different regions of microstructures, is often considerable the source of failure in a welded joints. During the welding thermal cycle, base metal close to the fusion area will transform to austenite, martensite ferrite, and bainite, depending on the cooling rate and steel composition. These different phase microstructure correspond to different mechanical properties. Since high temperatures are necessary during welding, a weld will exhibit a coarse crystal structure in contrast to the parent metal around it, which is generally in the wrought condition, some elements such as chromium in steel, accelerate grain growth, so that in these cases an enlarge grain may also be expected in the metal near to the weld [Lancaster, 1993].

A profile curve of the maximum temperatures reached at various location within the heat-affected zone of a weldment from 0.3 percent carbon steel may be correlated with the iron – iron Carbide phase diagram as shown below in Figure 2.1

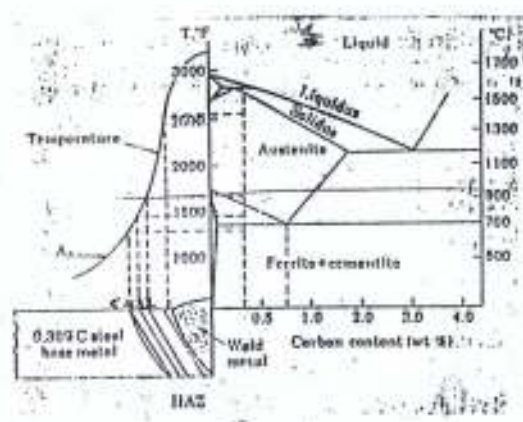


Figure 2.1: Profile curve of the maximum temperatures reached at various location within the heat –affected zone of a weldment. Source: American Welding Society

The structures found in the various regions of the heat-affected zone may be analyzed as follows.

- **POINT 1:** The steel in this zone has been heated to excess of 1301°C , so that austenite grains have grown to a large size.
- **POINT 2:** The alloy in this zone has been heated to 968°C and is fully austenized but there has been little grain growth.
- **POINT 3:** The steel in this region has been heated far enough above the austenite transformation temperature so that the austenite has achieved a homogenous structure.
- **POINT 4:** The metal in this region has been heated to 745°C . Only part of the alloy has changed to austenite, so that the resultant mixture of structure has poor mechanical properties.
- **POINT 5:** No change has occurred at this point, which represents base metal outside the heat-affected zone.

Higgin, (1993) investigated the changes that take place in the heat-affected zone with steel containing more than 0.3% carbon. His investigation revealed the presence of three separate sections in the heat-affected zone.

- (1) That part of the zone in which the temperature exceeds the upper critical level will undergo complete transformation to austenite grain size. On cooling the austenite grain size will be reflected in that of the ferrite – pearlite structure produced so that very near the weld a Widmanstätten structure may be present.
- (2) The region in which the temperature reached was between the upper critical (AC_3) and lower critical AC_1 or 730°C) will transform only partially. Pearlite will transform to austenite containing 0.8% carbon and then begin to absorb primary ferrite as the

temperature exceeds 723°C . However, subsequent cooling may be rapid enough, (this austenite being near-eutectoid composition) to form martensite so that a brittle region is produced in that part of the heat-affected zone).

- (3) In that section of heat-affected zone, where the temperature exceeds that of recrystallisation (about 500°C) the original cold-worked structure of the plate will tend to recrystallized without undergoing any phase change. He concluded that changes other than simple recrystallisation occur in the heat-affected zone.

Factors affecting heat-affected zone (HAZ) toughness are the nature of the weld thermal cycle, grain coarsening temperature, transformation characteristics, alloy content and non-metal content. Low heat input rate processes which give relatively high cooling rates, generate a fine-grained heat affected zone and less embrittlement in low carbon. In more hardenable steel (including carbon –manganese steels) this effect may be offset by the formation bainite or martensite microstructure. The high the carbon content, the more brittle the transformed structure. High carbon twinned martensite is the most brittle of the structure formed in the heat affected zone of steel and is particularly susceptible to hydrogen –induced cold cracking.

Hardness of the heat-affected zone is a measure of the tensile strength of the steel and, for any given alloy type, gives an indication of the degree of embrittlement. For carbon-manganese and some low-alloy high tensile steels, a hardness of over 350 BHN in the heat-affected zone would be considered excessive, indicating a susceptibility to cracking. Conversely, the heat-affected zone of nickel alloy steels may be adequately tough with a hardness of 400 BHN [Lancaster, 1993].

Rajanna et al, (2001) carried out a series of Charpy tests to examine the fracture toughness for the base, weld areas as well as the HAZ, and correlated the fracture behaviour to the grain size.

Sunghak et al, (1993) carried out fracture toughness analysis of heat-affected zones (HAZs) in high-strength low-alloy steel welds by detailed analysis of fracture processes, such as void initiation and propagation at sharp crack tip. The material tested is a 44-mm-thick plate of normalized steel. This is an HSLA steel with a tensile strength of about 500 MPa whose chemical composition is 0.3C, 0.415-1.43Mn, 0.019P, 0.015-0.065Al, 0.3Cu, 0.22Ni, 0.017Ti, 0.02Nb. And *in situ* fracture observation and parallel fracture initiation modeling are used. In order to explain weld joint performance, tensile and plain strain fracture toughness were conducted for simulated coarse-grained HAZ microstructures. The micro-mechanisms of fracture process involved in void and microcrack formation are identified by *in situ* scanning electron microscopy (SEM) fracture observations and void initiation study. The fracture toughness results are also interpreted using simple fracture initiation models founded on the basic assumption that a crack initiates at a certain critical strain or stress developed over some microstructural significant distance. Tensile test results conducted over the test temperature range of -80°C to -10°C for the simulated ICRCG HAZ and the metal showed that the tensile and yield strength rise steadily with decreasing test temperature. Figure 2.6 (a) the general trends are substantially higher strengths for the ICRCG HAZ, especially tensile strength.

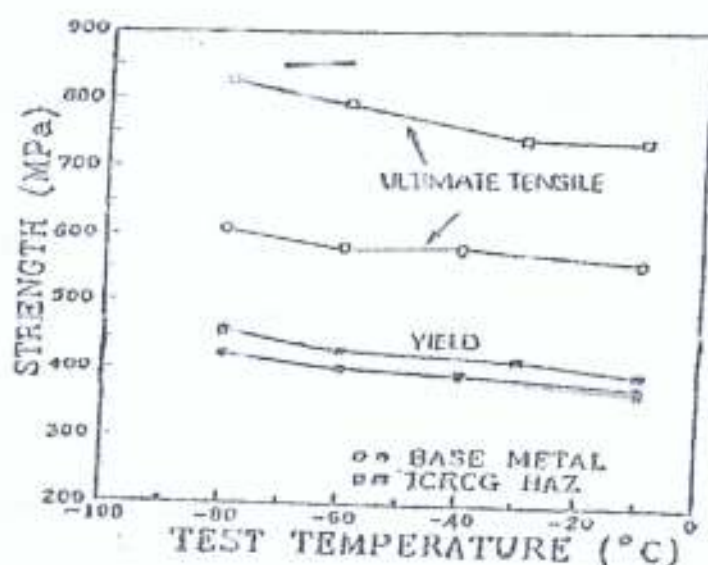


Fig. 2.2: Ultimate tensile strength and yield strength of the simulated ICRCG HAZ and the base metal vs test temperature

Martensite Island play an important role in reducing the toughness coarse-grained HAZs through their effect on reduction of critical distance between void initiation sites. It concludes, that the formation of martensite Islands should be minimized during welding to improve the fracture toughness of the welded joints. The performance of welded structure is usually limited by failure initiation within the heat-affected zone (HAZ) of the base plate, especially the coarse-grained heat-affected zone (CG HAZ) adjacent to the weld metal. Even though some phases in the heat-affected zone HAZ show their brittleness and sensitivity to the micro cracks, one cannot assume that the initiation of failure for a weld joint without Pre-defect under uniform dynamic loading is always in the HAZ. The source and mechanisms of failure for a weld joint without pre-defects under uniform dynamic loading still need to be investigated and the strain rate sensitivity of fracture mechanism cannot be excluded.

The heat input rate is one of the most important variables in fusion welding. It governs heating rates, cooling rates and weld pool size. The higher the heat input rate the lower the

cooling rate and larger the weld pool. Low input rate process which give embrittlement in low – carbon steel

According to Lancaster, (1993) there is an inverse relationship between weld pool size and cooling rate. The heat input rate q per unit length of weld is given by the equation.

$$\frac{\eta VI}{V} = \frac{q}{v} \dots\dots\dots 2.3$$

where V represent arc voltage

I represent current

η proportion of arc energy that is transferred as heat to work piece

v welding speed

The other metallurgical feature that is directly affected by heat input rate is the grain size in the heat – affected zone and in the weld metal. Grains in the solidifying weld metal grow coherently with grains in the solid metal at the fusion boundary. Therefore, the longer the time spent above the grain coarsening temperature of the alloy in question, the coarser the structure in the heat – affected zone and in the weld metal. In face-centred cubic metals (FCC) and alloys, coarse grain size may not have much effect on the weld behaviour, but in ferritic steels, various properties, particularly notch ductility, deteriorate as the grain size increases. A high heat input rate process will, in general, give a longer thermal cycle and tend to generate a coarser structure. In steel welding it is therefore necessary to seek a heat input rate that gives the optimum combination of grain size and cooling rate

Correlation of the weld thermal cycle and the microstructure with fracture toughness is extremely important for understanding fracture in welded joints, but much remains to be learned concerning the actual microfracture process that occur at a crack tip. Researches have revealed that fracture properties of the intercritically reheated coarse-grained heat-affected zone (ICRCG

HAZ) i.e. local brittle zone (LBZ), are strongly influenced by metallurgical factors such as high-carbon martensite Islands. This is usually characterized by a trough in the plot of Charpy impact energy as a function of second-peak reheating temperature (T_p^*). Charpy energy was found to decrease monotonically with an increasing amount of martensite Islands, confirming that the martensite Island is the major microstructural factor controlling the HAZ toughness. Since this embrittlement problem in multipass weld joints limits the service temperature of high-strength low alloy (HSLA) steels [Sunghak et al, 1993].

2.10 HARDNESS DISTRIBUTION IN WELD METAL

The contour of Rockwell hardness along the sectional direction of a welded plate is as shown in Figure 2.3, it is easy to see that the hardness outside the welded area (base plate) is higher than that in the inside area (weld metal). Inside the welded region, the hardness is approximately HRC19, while the base metal has hardness HRC 25. The point close to the boundary, between the weld steel and the HAZ, shows higher hardness than the weld center area.



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Fig. 2.3: Rockwell hardness contour map of the measurement in C scale

Q.xuel et a, (2002) investigated the mechanical response of welded high – alloy (HSLA) 100 steels by both quasi – static ($103S^{-1}$) and high strain rate($103S^{-1}$) tension loadings at ambient and low temperature. The three quasi – static stress curve in Figure 2.4 represents mechanical responses of the materials from the base metal, the weld metal and their boundary with the HAZ in the center of the specimen.

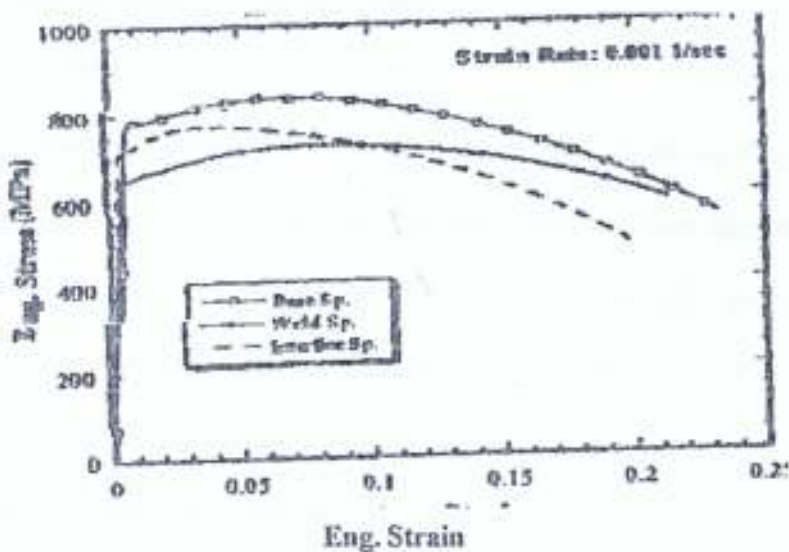


Fig. 2.4: Comparison of Quasi-Static Responses At Ambient Temperature

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He reported that the base metal shows the highest yield stress, while the weld metal gives the lowest one. The tensile test does not however capture the small fluctuations in yield stress.

All interface specimen under different strain-rate temperature show that necking happened inside the welding area, figure2.5 is a sectional view of an interface specimen

All interface specimens under different strain – rate and temperatures shows that the necking happened inside the welding area Figure 2.5 is a sectional view of an interface specimen . The necking always happens in the weld metal region.



Fig. 2.5: Failure configuration of the interface specimen. Q. xuel et al, 2003.

The temperature effect is another important factor in the dynamic tension tests, three testing temperatures, the ambient temperature (298k), 190k and 77k were chosen for investigating the low-temperature behaviour, of the weld element. Result shows that the lower the temperature, the more brittle the material appears. For the liquid nitrogen temperature, the strength and the yielding stress increase about 35 and 53.1%, compared with the quasi-static data. The total elongation under the same loading duration decreases to about 60% of that the ambient temperature for the 190k, which is close to the real low-temperature environmental in polar area for naval ship, the strength and yield stresses increases about 12 and 23.8 while the elongation under the same condition decreases by 5%.

2.11 ELECTRODE

The composition of the core of wire depends upon the metal of the weld, for example, to weld mild steel core wire of similar composition will be prepared, in order to get a homogenous weld joint. The size and diameter of the core wire will depend upon the amount of the weld to be deposited and on the type joint of the joint or the gap to be bridge between the two plate to be welded [Lancaster, 1993].

2.11.1 ELECTRODE TYPES

2.11.1.1 NON -CONSUMABLE OR REFRACTORY ELECTRODES

They are made up of high melting point material like carbon (Mp 6700° F), pure tungsten (MP6150F) or alloy tungsten . These do not melt during welding; they maintain which melt the base metals (as in TIG and carbon arc welding). The electrode length goes on decreasing with the passage of time, because of vaporization and oxidation of the electrode material during welding example of non consumable electrode are copper coated carbon or graphite electrodes.

2.11.1.2 CONSUMABLE ELECTRODE

These electrodes possess higher thermal efficiency (about 85%) compared to that of non-consumable electrode welding arrangement (about 55%) consumable electrode; bare electrodes, shielded electrodes, medium coated electrodes, and heavily coat electrode covered electrode (slightly ,medium and heavily coated electrodes)produce good weld appearance , weld metal properties and defect free joint, covered electrode may be categorized as follows;

- 1 Electrode which have cellulose content impart deep penetration and increased electrode burn up rate , welding may be carried out in all positions . Example of this are AWS 6010, 6011, 7010, 7011, Bs E110 IS (814-1970)E,100,22A, M100264.
- 2 Rutile (titania) electrode give a smooth and quiet arc , negligible spatter, and easily removable slag . the electrode work both on AC and DC. Other ingredient position. Examples of such electrodes are AWS E6012, 6013, 7013, MB 17275, E31432 C
- 3 Electrode containing iron oxide produce fluid good weld appearance and mechanical properties. This electrode can be operated on Ac and DC Slag is easily detachable. Because of slag fluidity, they are general employed for flat welding. Example include E31432 C

- 4 Electrode with iron power content have stable arc and large metal deposition rates, example of these electrode of these electrode are E6027, 6024, BS922P is M722
- 5 Low hydrogen electrode contain low moisture content fluorspar limestone e.t.c they produce weld highly resistance to cracking and for a maximum notch ductility. Such electrode need some heating before used they work in all position and they can be operated on AC and DCRP. Example of these electrodes are AW E6016, 7016, BS E 616H, 15M616489H, e.t.c (Gourd, 1993)

2.12 SELECTION OF ELECTRODE

- Selecting the right kind of electrode(s) for a particular application is of utmost important in welding if desired properties in the weld joint are to be achieved. Gourd (1993), shows that the choice of electrodes depend on the following; chemical composition of base metal, thickness of the work pieces, nature of the electrode coating, type of joint, type of power source, and type of polarity.

CHAPTER THREE



MATERIALS AND METHOD

MATERIALS

The material studied was low-carbon steel in the form of plate of a 10-mm thickness obtained from metal scrap market at Owode Onirin in Ikorodu, Lagos, Lagos State. The chemical analysis of the steel was conducted at Nigerian Foundries Limited, Ilupeju, Lagos, Lagos State, by optical emission spectrometry using an AR4 30 metal analyzer. The chemical composition is listed in Table 3 below.

TABLE 3: *Chemical Composition of the Experimental Steel plate*

Element	Fe	C	Si	Mn	S	Cu	Al	Co	Ti
% wt	98.89	0.23	0.31	0.48	0.028	0.010	0.005	0.045	0.005

METHOD

The as-received steel plate was cut into forty five (45) samples of dimensions 10 x 10 x 240mm with hacksaw at the fittings laboratory of the Department of Mechanical Engineering, University of Lagos. The samples (45) were classified into three groups designated by K.L.M with each groups having fifteen (15) samples.

HEAT – TREATMENT

The samples were heat treated at the Metallurgical laboratory of the Department of Metallurgical Engineering, University of Lagos. Group “k” samples were annealed by heating to 920°C soaked for 45 minutes and furnace cooled to obtain *annealed structure*. Samples in-group ‘L’ were heated to 920°C, allowed to soak for 30 minutes and cooled in air to obtain *normalized structure*. And group ‘M’ samples were left untreated as *as – received structure*.

3.4 DEFORMATION

All samples in the groups, K, L, M were rolled on a laboratory mill (Work roll diameter of 180mm) from the initial state of deformation at zero (0%) percent reduction to five different degrees of deformation (percent reduction) of 5%, 10%, 20%, 30% and 40%. Rolling was carried out in the machine/foundry workshop of the Department of Metallurgical/Materials Engineering, University of Lagos, Lagos State.

3.5 WELDING

Welding was carried out at the welding and fabrication workshop of the Mechanical Engineering Department, University of Lagos, with an Esab welding machine. Five (5) samples each from groups K,L,M were cut into two equal halves with hacksaw, each half was joined to the corresponding one by welding (butt joint) to give 'welded samples; The weld metal was deposited by multi-pass electric arc welding at a voltage of 220V and a current of 180A (low heat input) using commercially available multi-purpose electrode/general purpose electrode (E-6013).

All samples from the annealed, normalized and as-received structures were heat-treated by stress-relief annealing at 650⁰C to remove the effect of internal stresses imposed on them during rolling and welding.

3.6 METALLOGRAPHY

Optical microscopy studies were conducted on the 'test' samples at the Metallurgical Laboratory of the Department of Metallurgical and Materials Engineering Obafemi Awolowo University, Ile-Ife, Osun State. Metallography samples of dimensions 10 x 10 x 30mm were cut with hacksaw from the non welded (base plate) and welded (weld metal) of all the structures,

MECHANICAL TESTING

Samples of dimensions 10 x 10 x 210mm that were left from initial starting dimensions 10 x 10 x 240mm were machined to final dimensions for hardness, tensile and impact tests. Hardness tests were carried out on the samples at the stress analysis Laboratory of the Department of Mechanical Engineering University of Lagos. Samples of dimension 10 x 10 x 70mm were cut with hacksaw from all the structures (annealed, normalized and as-received). Hardness measurements were carried out across the non welded parts (base plates) and welded parts (weld metal) using a Rockwell hardness direct indentation machine with Rockwell B scale (tip load 100kg), an average of seven (7) indented field at the base plate and weld metal were reported for all samples at ambient temperature (25°C).

For tensile test measurements, tensile specimen of a 25-mm gauge length and a 6.25-mm gauge diameter were used in the present investigation for the non welded (base plate) and welded (weld metal), machining was done with lath machine at SOWAG Engineering company, Akure, Ondo State. The tensile tests were carried out at the metallurgical laboratory of the Metallurgical and Materials Engineering Department, Federal University of Technology, Akure, with a Monsato type W-Tensometer at a crosshead speed of 1mm/min and at ambient temperature (25°C)

Standard charpy samples with dimension 10 x 10 x 60mm and a 2-mm V-notch were prepared for non welded parts (base plate) and welded parts (weld metal) at the fittings laboratory of the Department of Mechanical Engineering, University of Lagos. Thereafter, impact energy (notch toughness) test were carried out on the various structures at ambient

impact energy (notch toughness) test were carried out on the various structures at ambient temperature (25°C) at the stress analysis laboratory of mechanical engineering department, University of Lagos.

CHAPTER FOUR

4.0 RESULTS AND DISCUSSION

4.1 RESULTS

4.1.1 MICROSTRUCTURES



Optical microscopy studies of the non-deformed as-received low-carbon steel plate revealed predominantly pearlite phase in a matrix of ferrite Plate 4.1 (a). The dark areas correspond to pearlite and the light areas depict the ferrite phase. Extremely fine precipitates are seen dispersed as dark spots in the matrix, after annealing and normalizing heat-treatments at 920°C with soaking time of 45 minutes and 30 minutes and cooled in furnace and air respectively, the microstructures of the non deformed annealed and normalized structures are as given in Plates 4.1(b) and (c) respectively, they revealed mixed phases of ferrite (white patches) and pearlite (dark) similar to the as-received condition (as-received structure), also observed were dispersed fine precipitates as dark spots oriented in all direction in the matrix, steel (4.1b) in normalized condition (normalized structure) Plate 4.1(c) revealed microstructure with more fraction volume of pearlite and less ferrite as compared to steel in annealed condition (annealed structure).



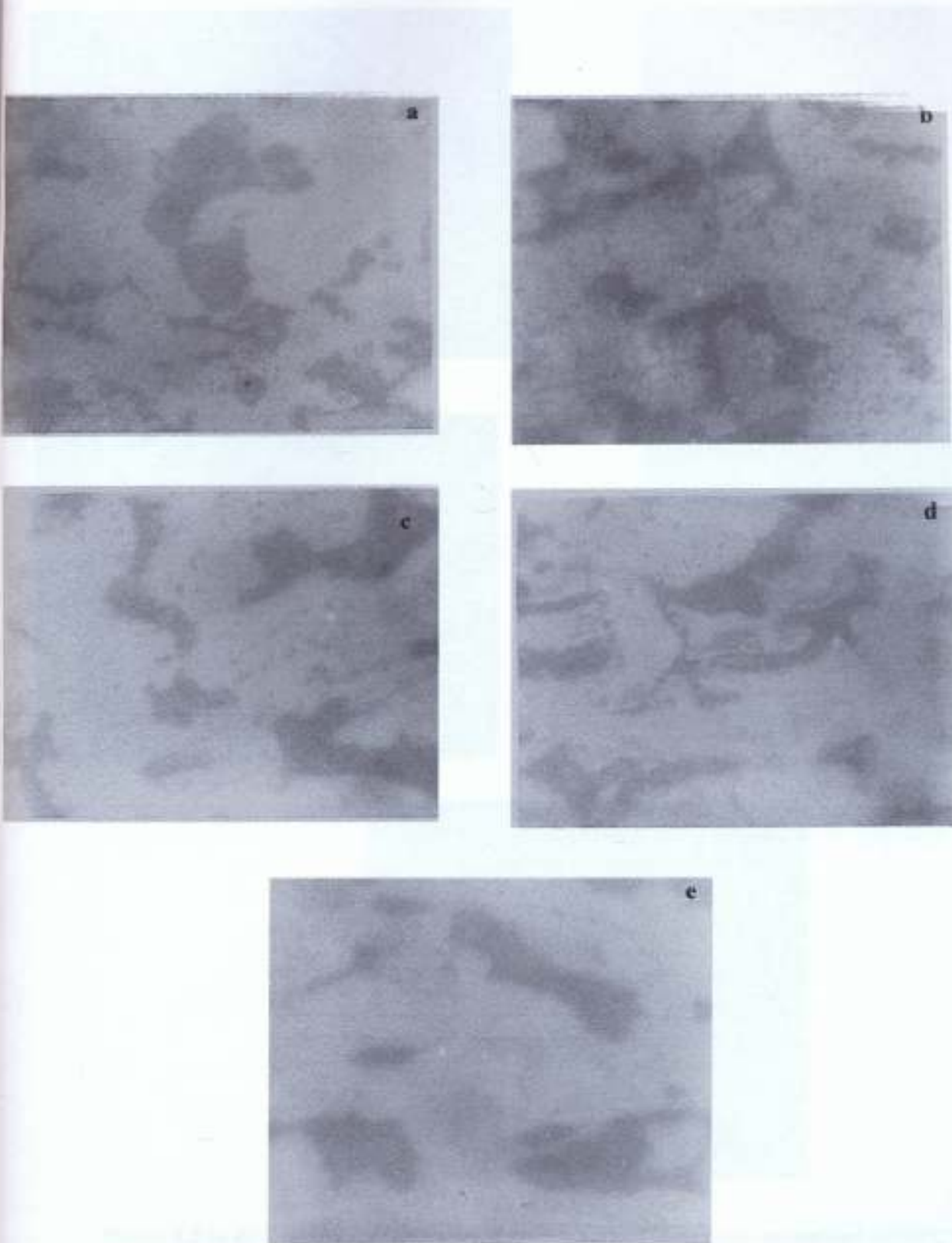
Plates . 4.1(a-c): Optical microstructures of the non – deformed base plate (a) in as-received condition (b) annealed at 920°C (c) normalized at 920°C showing pearlite (dark areas) ferrite (light areas) with precipitates (dots). Etched in 2pct natal, 400x

Fineness of grains were also observed in the various structures, only marginal difference in grain size between the non deformed as – received and annealed structures were revealed Plates 4.1(a) and (b) respectively, with the as-received structure having slightly elongated grains and the annealed structure roughly better large equiaxed grains and smooth boundaries, pearlite phase is better dispersed in ferrite matrix and precipitates are bigger.

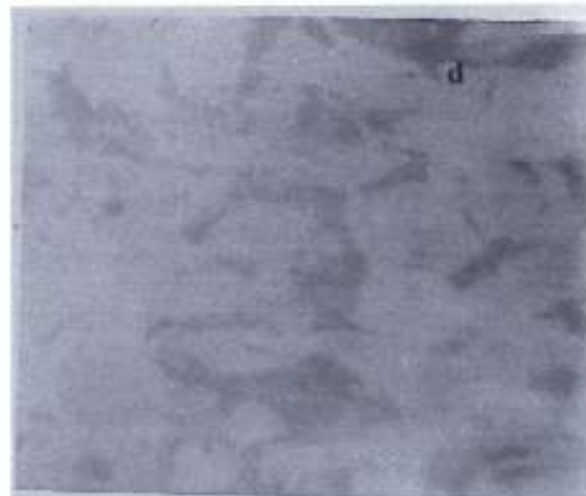
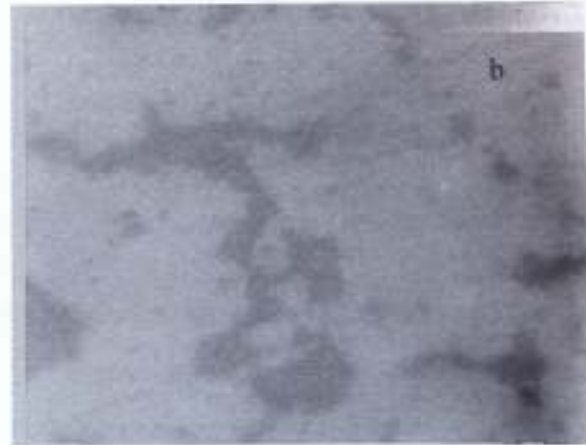
Optical microstructures of the annealed structure cold – worked at 5%, 10%, 20%, 30%, and 40% reduction respectively from initial state of 0% deformation are as shown in Plates 4.2 (a) through (e). Distorted pearlite phase in ferrite matrix is seen.

Optical microstructures of the normalized structure, cold – worked at 5% 10% 20% 30% and 40% from initial state of 0% deformation are shown respectively in Plates 4.3(a) through (e). The micrographs also revealed predominantly mixed phases of ferrite and pearlite as in annealed structure, but with relative more fraction volume of distorted pearlite in ferrite matrix.

Optical microstructures of the as – received structure deformed to 5, 10, 20, 30, and 40 percent from initial condition of 0% deformation is provided in Plates 4.4 (a) through (e), observed were distorted pearlite in ferrite matrix as in the cases of annealed and normalized structures. However, for equal percent reduction lattice distortion and grain elongation are less severe compared to the normalized structure, while only marginal difference in grain elongation exists between the as- received and annealed structures



Plates . 4.2 (a-e): Optical microstructures of the non welded (base plate) Annealed at 920°C for 45 min and cold rolled at ambient temperature (25°C) in the ranges 5%, 10%, 20%, 30 % and 40% respectively showing elongated pearlite (dark areas) in ferrite matrix (light areas). Etched in 2 pct nital 400X

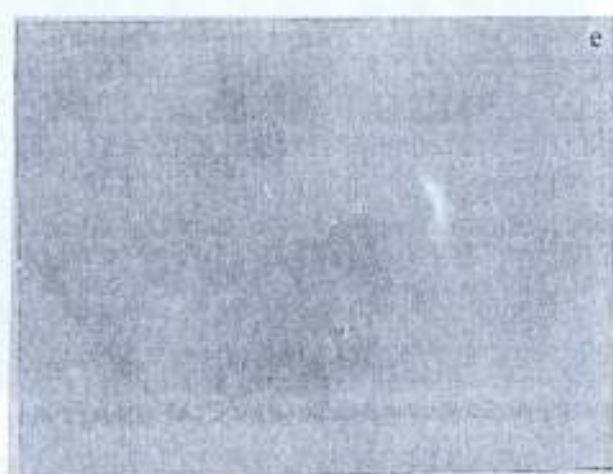
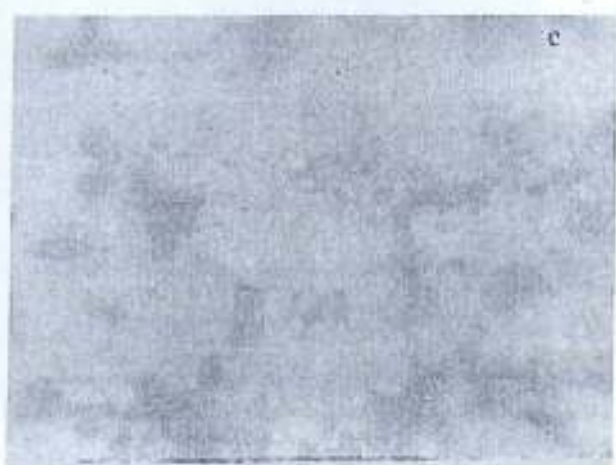


Plates: 4.3 (a-e): Optical microstructures of the non welded (base plate) normalized at 920°C for 30 mins, cold rolled at ambient temperature (25°C) in the ranges 5%, 10%, 20%, 30% and 40% respectively showing more fraction volume of elongated pearlite (dark areas) in ferrite matrix. Etched in 2 pct natal



Plates 4.4 (a-e): Optical microstructures of non welded (base plate) in as-received condition, cold rolled at ambient temperature (25°C) in the ranges 5%, 10%, 20%, 30% and 40% respectively, respectively (showing elongated pearlite (dark areas) in ferrite matrix. Etched in 2 pct natal 400X

Microstructures of the welded parts (weld metal) annealed, normalized and as-received structures at the various percent reduction are as shown in Plates 4.5 – 4.7(a-e) respectively. The optical micrographs showed rather equiaxed grains, especially at high percent reduction, but at low percent reduction slight grain elongation comparable to that of the non welded parts (base plate) structures are observed. For the normalized structure Plates 4.6(a - e) at every percent reduction, growth rate of the recrystallized grains was observed to be slower, also observed were relatively more equiaxed grain as compared to the welded annealed and as- received structures. However, the welded as-received structure revealed grains with smaller size as compared to the welded annealed structure.



Plates 4.5 (a – e): Optical microstructures of welded steel plate annealed at 920°C for 45mins cold rolled at ambient temperature (25°C) in the ranges 5%, 10 %, 20%, 30% and 40% respectively showing pearlite (dark) in ferrite (light) matrix. Etched in 2 pct natal, 400x



Plates 4.6 (a- e): Optical microstructures of welded steel plate normalized at 920°C for 30 mins, cold rolled at ambient temperature 25°C in the ranges 5%, 10%, 20%, 30% and 49% respectively showing pearlite (dark) in ferrite (white) matrix. Etched in 2pct natal. 400x



Plates 4.7(a-e): Optical microstructures of the welded steel in as-received condition cold rolled at ambient temperature (25°C) in the ranges 5%, 10%, 20%, 30% and 40% respectively showing pearlite (dark) in ferrite (white) matrix. Etched in 2pct natal, 400x

4.1.2 MECHANICAL PROPERTIES

Hardness of the steel plate as a function of percent reduction carried out on the non welded and welded parts at ambient temperature (25°C) are as plotted in Figures 4.1(a)-(f), and the corresponding bar charts are as shown in appendix v, Tables 4.1 (a)-(c) in appendix i were used for the plots. Observation made showed that at all percent reductions non welded parts (base plate) of the annealed, normalised and as-received structures exhibited superior hardness over the corresponding welded part (weld metal), notice made also revealed that for both welded and non welded parts, hardness increases with increasing percent reduction with the normalized structure exhibiting superior hardness over the annealed and as – received structures

Yield strength (YS) of the steel plate as a function of percent reduction carried out on the non welded part (base plate) and welded part (weld metal) at ambient temperature (25°C) are as shown in Fig.4.2 (a)-(e), and the corresponding bar charts are as shown in appendix vi, Tables 4.2(a) –(c) in appendix ii were used for the plots, like in the case of hardness, observation revealed that for the various structures the non welded part exhibited superior yield over the corresponding welded part. Yield strength (YS) of non welded part of the annealed structure was observed to increase in increasing order as percent reduction increased, but at a relative low rate as compared to non welded parts of the as-received and normalized structures, at higher percent reduction an incoherent yield in the welded part of the as-received structure Fig. 4.2(b) was observed as compared to the welded parts of the normalised and the as-received structures with the welded part of th normalised structure having an improved yield at all corresponding percent reduction over that of the as-received structure.

Tensile strength (TS) of the steel plate with percent reduction carried out on the non welded part (base plate) and welded part weld metal at ambient temperature (25°C) are shown in

Figures 4.3 (a-e) and the corresponding bar charts are shown in appendix vii, Tables 4.3 (a)- (c) in appendix iii were used for the plots. For all the various structures, tensile strength (TS) of the non welded part and welded part were observed to increase with increasing percent reduction, with the non welded part exhibiting superior tensile over the welded part, unlike yield strength which showed improved values over tensile strength at low percent reduction, at higher percent reduction, the tensile strength of the various structures of non welded and welded parts was observed to be relatively higher than yield strength. Tensile strength of non welded part of the normalized structure was noticed to increase slightly at a decreasing rate and further at appreciable rate at higher percent reduction, while for the welded part of the as-received structure a relatively more consistent increase in tensile was noticed as deformation was increased from 0 to 40 percent.

Impact (notch toughness) energy of the steel plate as a function of percent reduction carried out on the non welded (base plate) and welded part (weld metal) at ambient temperature (25°C) are given in Figures 4.4 (a) – (e) and the corresponding bar charts are shown in appendix viii, Tables 4.4 (a)-(c) in appendix iv were used for the plots, at all percent reduction impact (notch toughness) energy of the non welded parts of the normalized, annealed and as-received structures were inferior as compared to that of the corresponding parts of the welded structures, this is a sharp contrast from hardness and tensile properties of the non welded and welded parts of the corresponding structures. Unlike in hardness and tensile, an interplay of impact energy between non welded parts of the annealed and as-received structures were observed at lower and higher percent reductions Fig.4.4 (a), and at about 7% reduction, an interplay between that of the annealed and normalised structures were observed as well Fig. 4.4 (b) with the annealed structure exhibiting superior impact over the normalised structure as percent reduction was increased

Relative low impact values were obtained with the non welded and welded parts of the as-received structure both at low and high percent reduction.

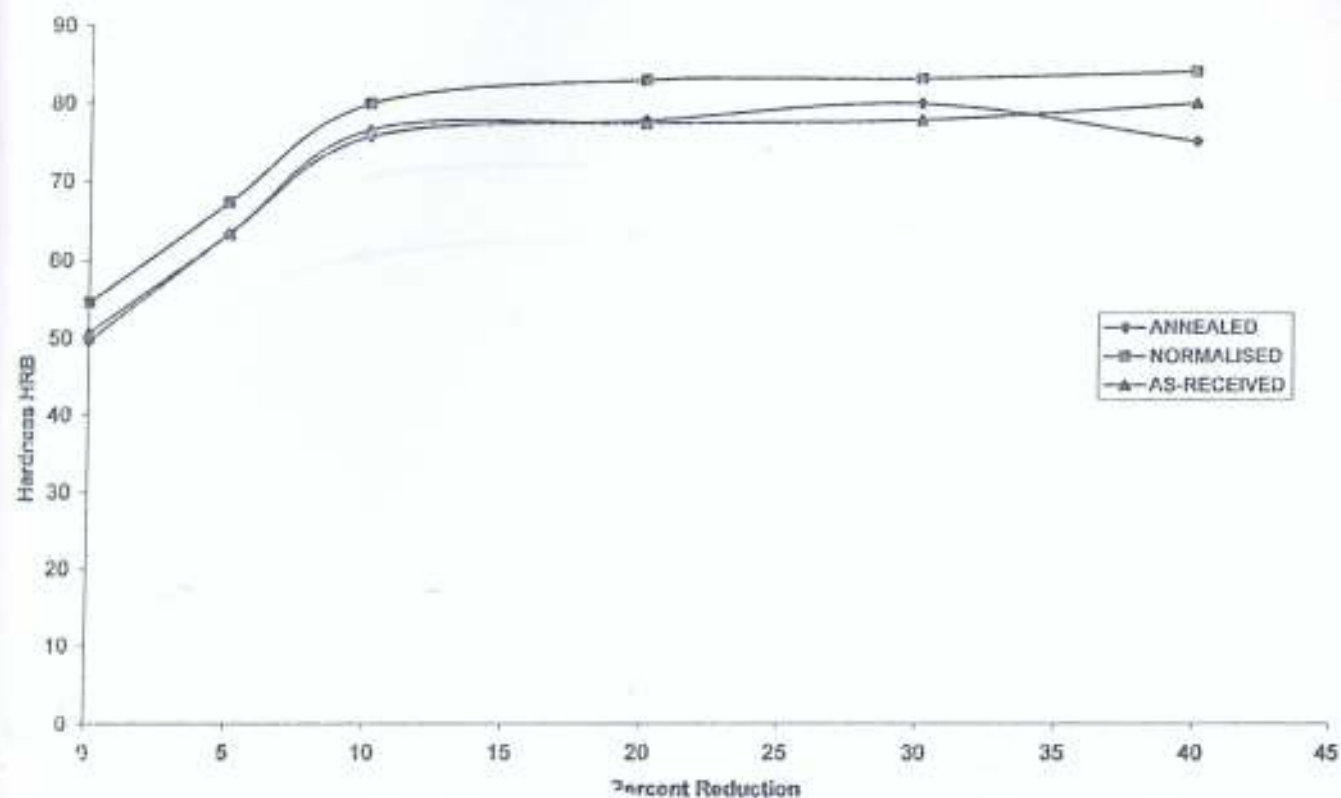


Fig. 4.1(a): Hardness of Non Welded (Base Plate) Annealed, Normalised and As Received Structures at Ambient Temperature (25°C).

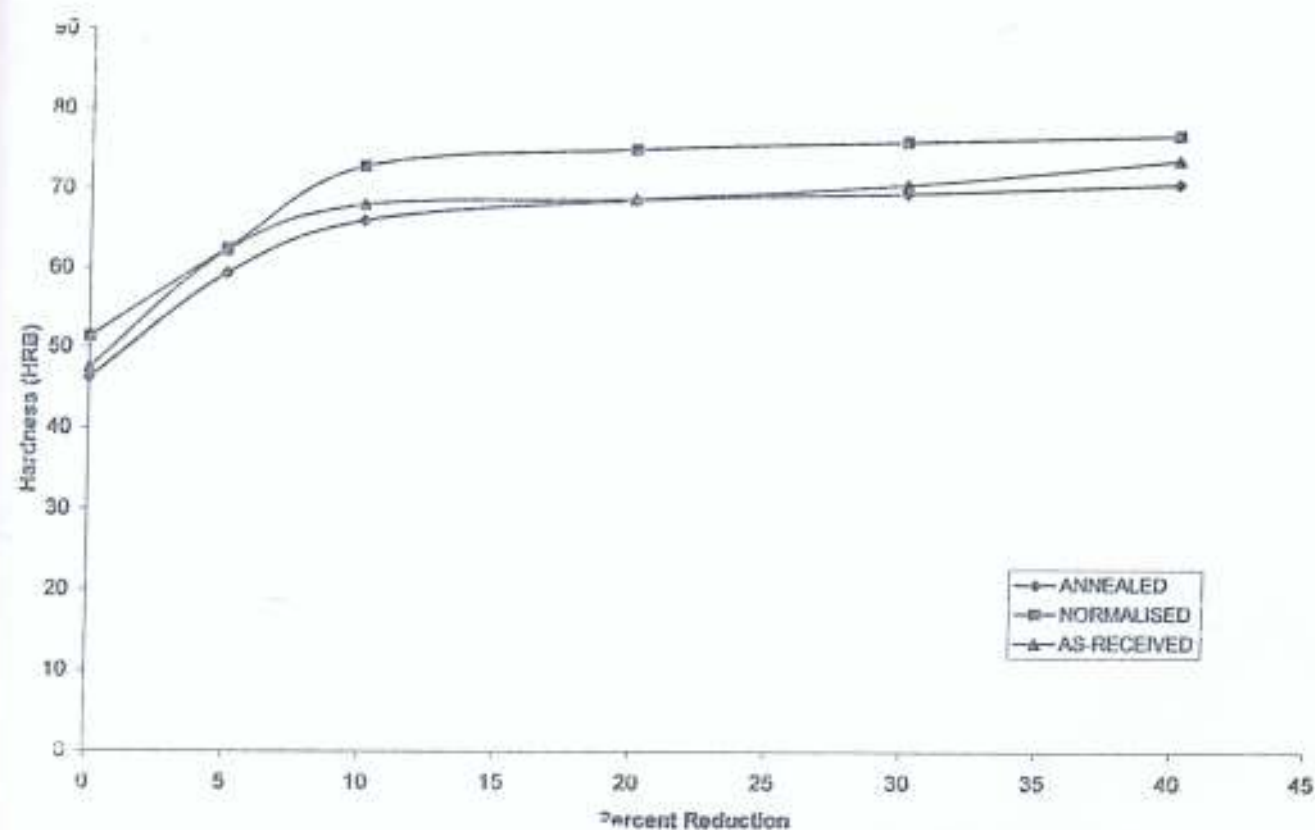


Fig. 4.1(b): Hardness of Welded (Weld Metal) Annealed, Normalised and As-Received Structures at Ambient Temperature (25°C)

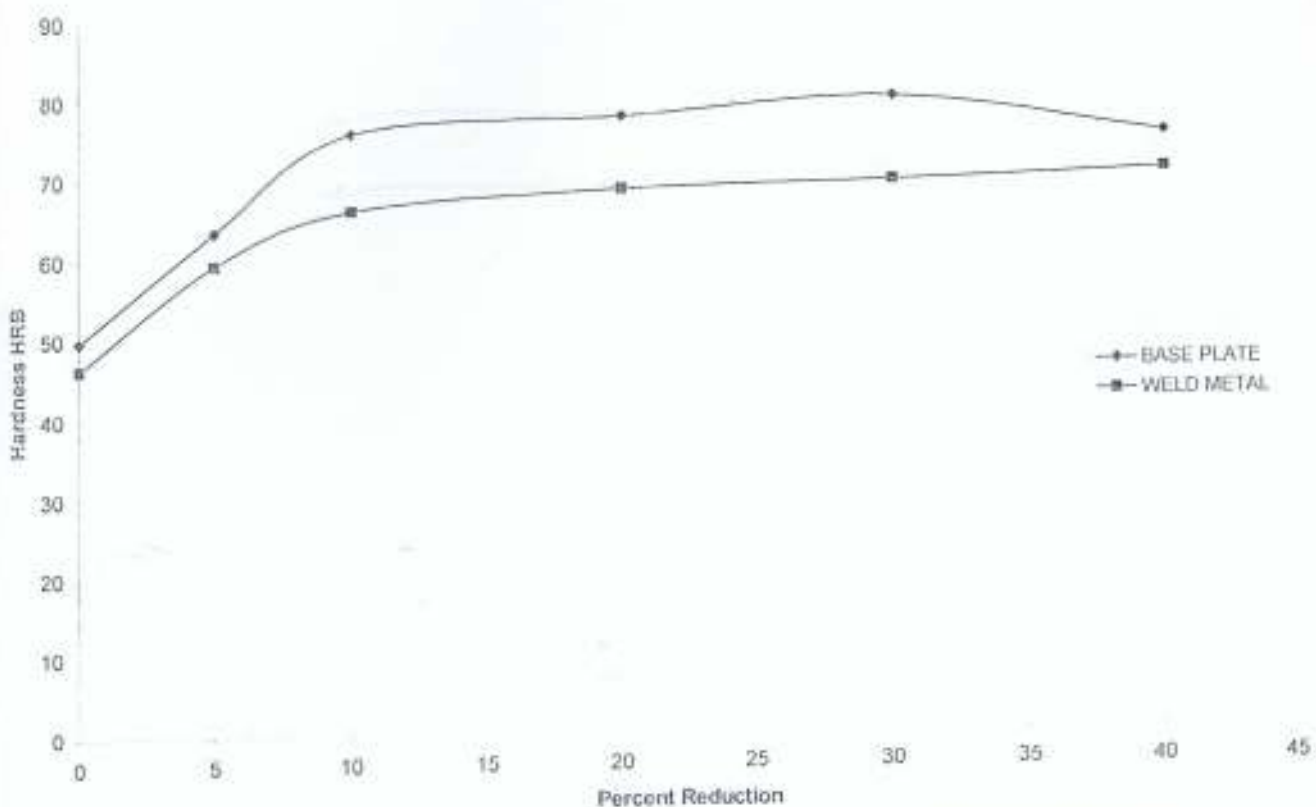


Fig. 4.1(c): Hardness of Non Welded (Base Plate) And Welded (Weld Metal) Annealed Structure at Ambient Temperature (25°C)

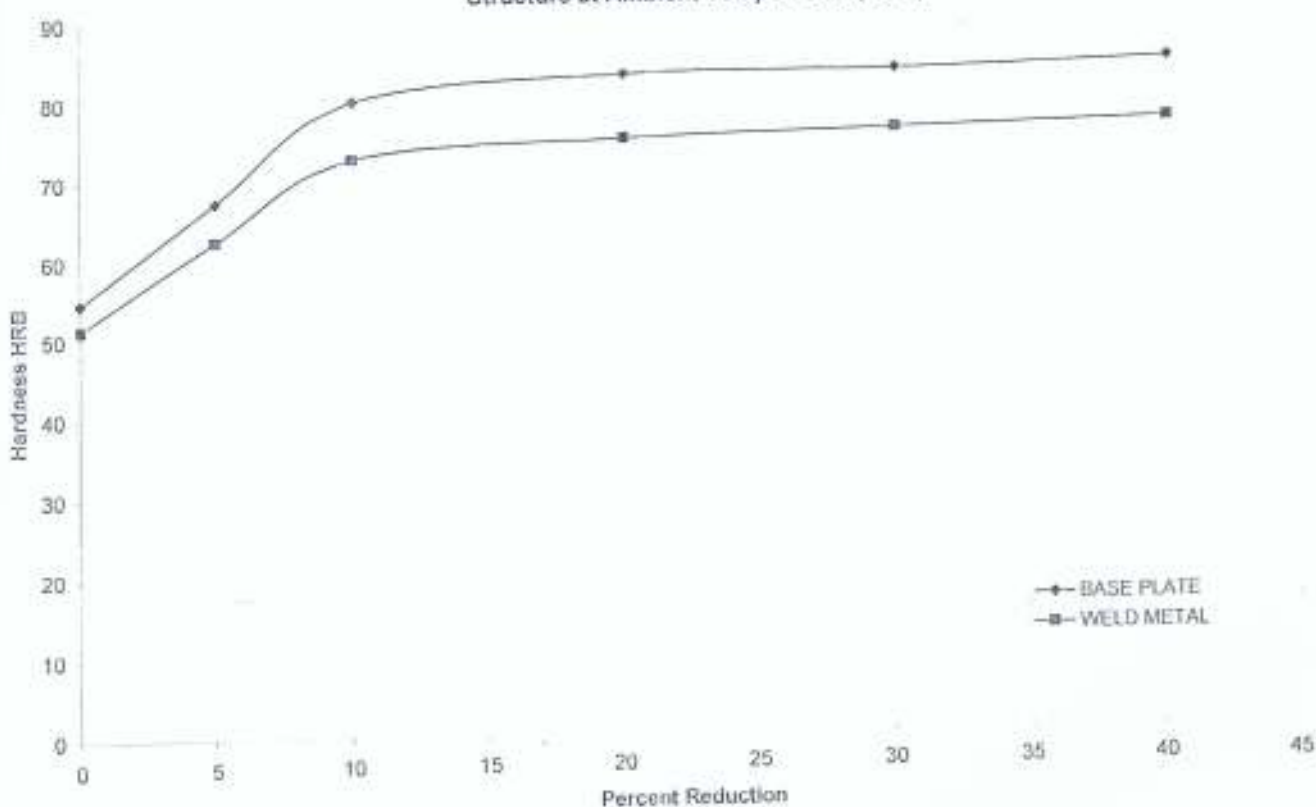


Fig. 4.1(d): Hardness of Non Welded (Base Plate) And Welded (Weld Metal) Normalised Structure at Ambient Temperature (25°C)

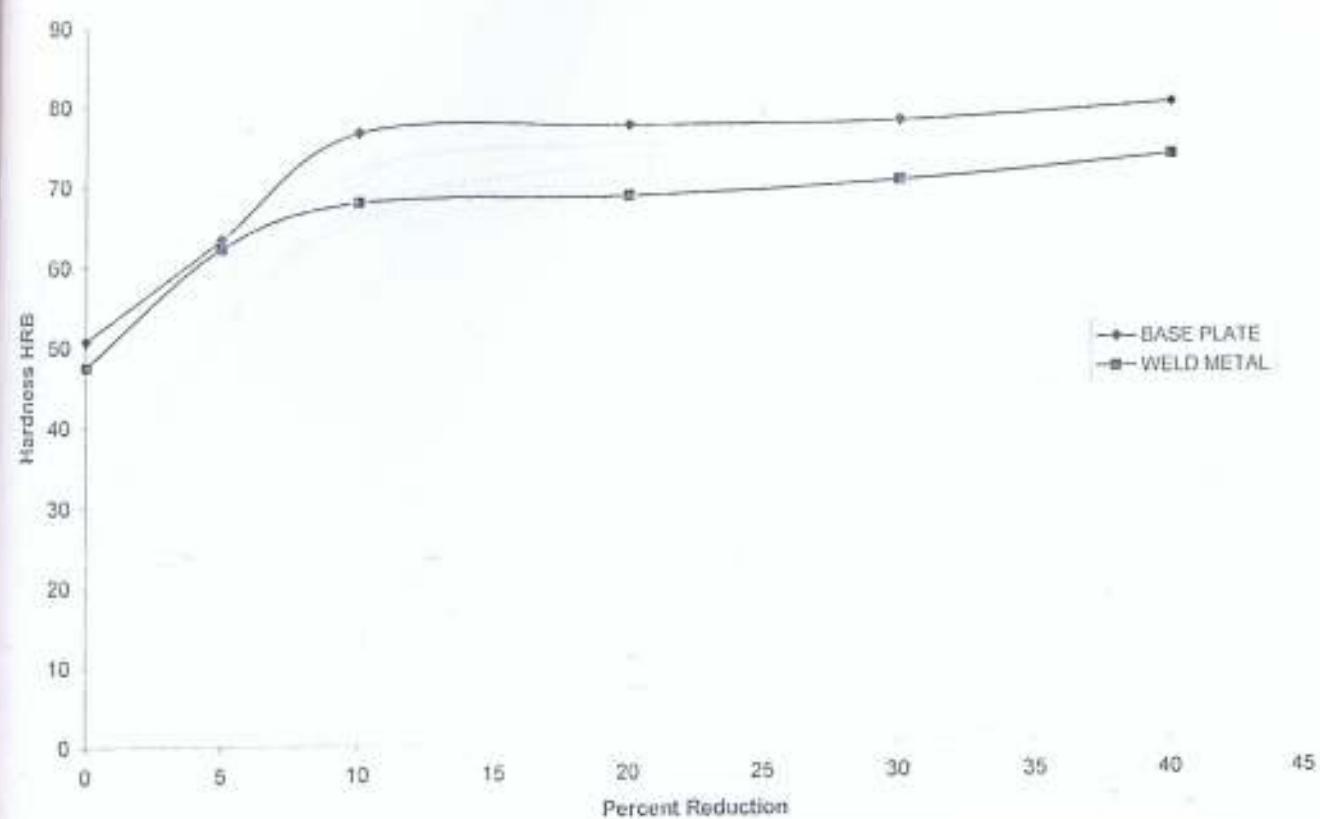


Fig. 4.1(e): Hardness of Non Welded (Base Plate) And Welded (Weld Metal) As-Received Structure at Ambient Temperature (25°C)

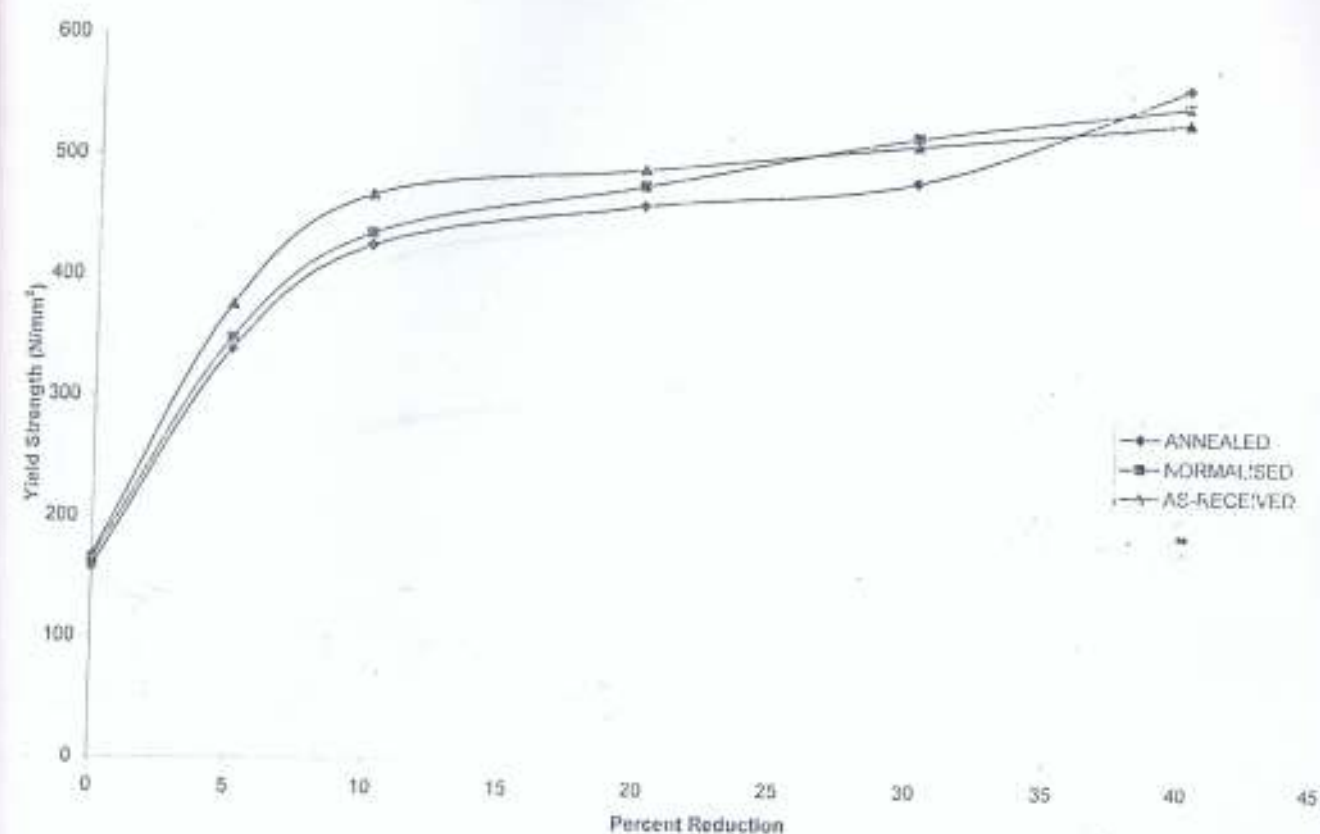


Fig. 4.2(a): Yield Strength of Non Welded (Base Plate) Annealed, Normalised and As-Received Structures at Ambient Temperature (25°C)

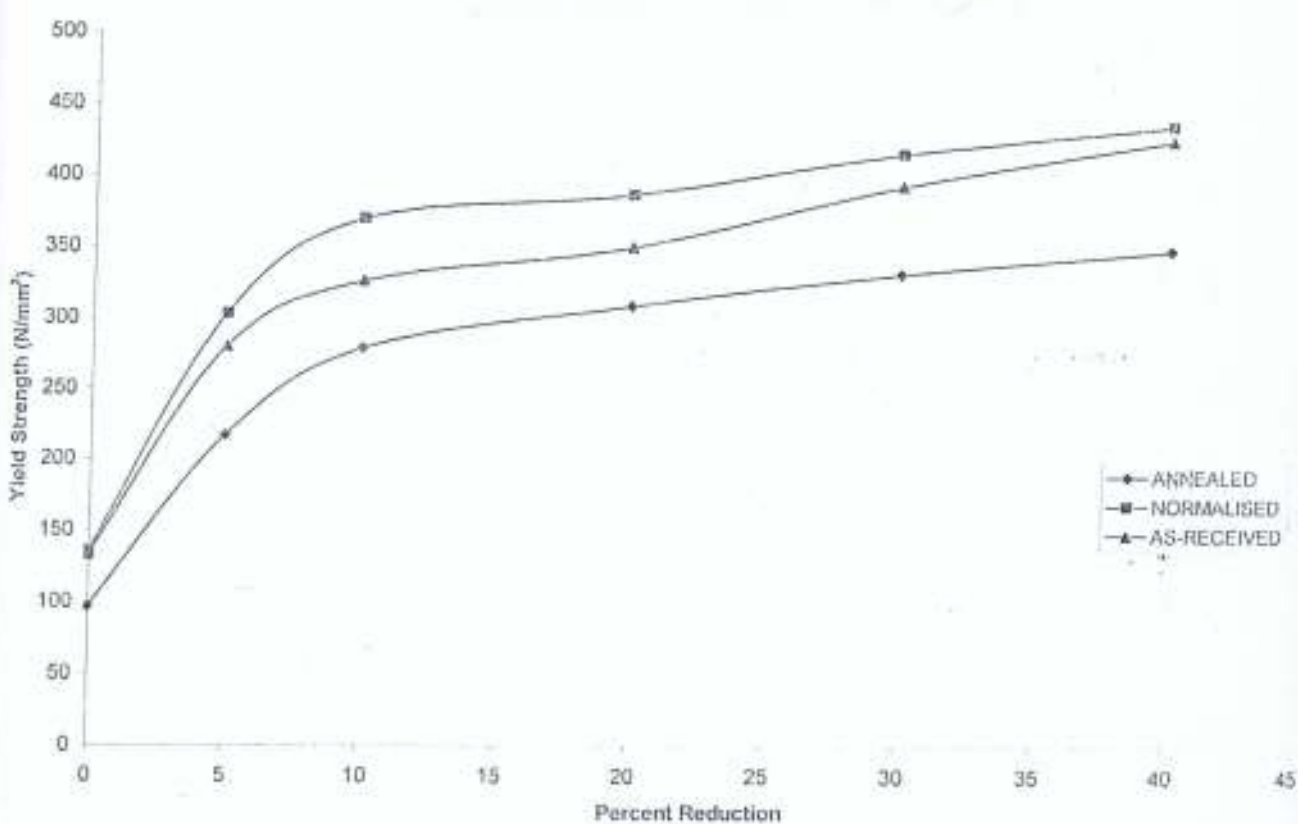


Fig. 4.2 (b): Yield Strength of Welded (Weld Metal) Annealed, Normalised and As-Received Structures at Ambient Temperature (25°C)

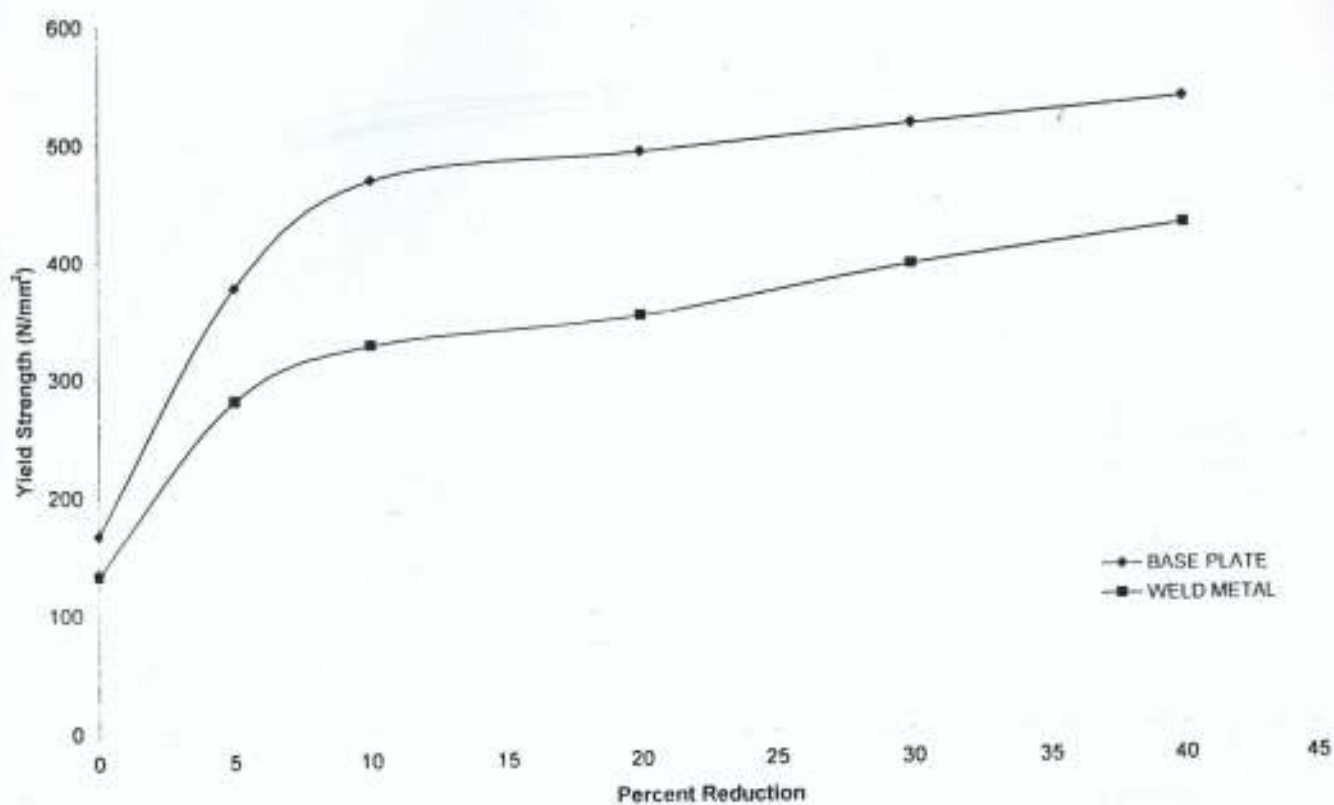


Fig: 4.2(e): Yield Strength of Non Welded (Base Plate) And Welded (Weld Metal) As-Received Structure at Ambient Temperature (25°C)

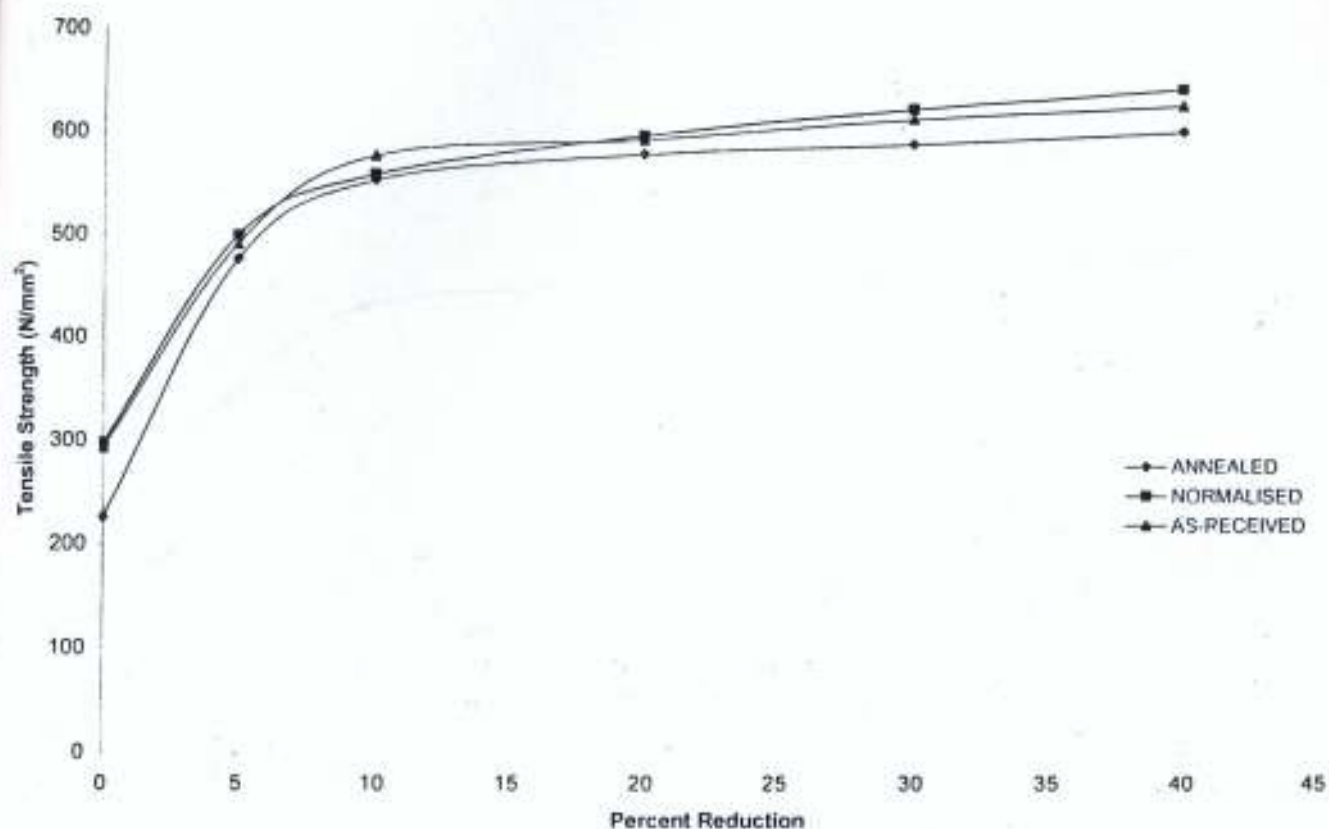


Fig: 4.3(a): Tensile Strength of Non Welded (Base Plate) Annealed, Normalised and As-Received Structures at Ambient Temperature (25°C)

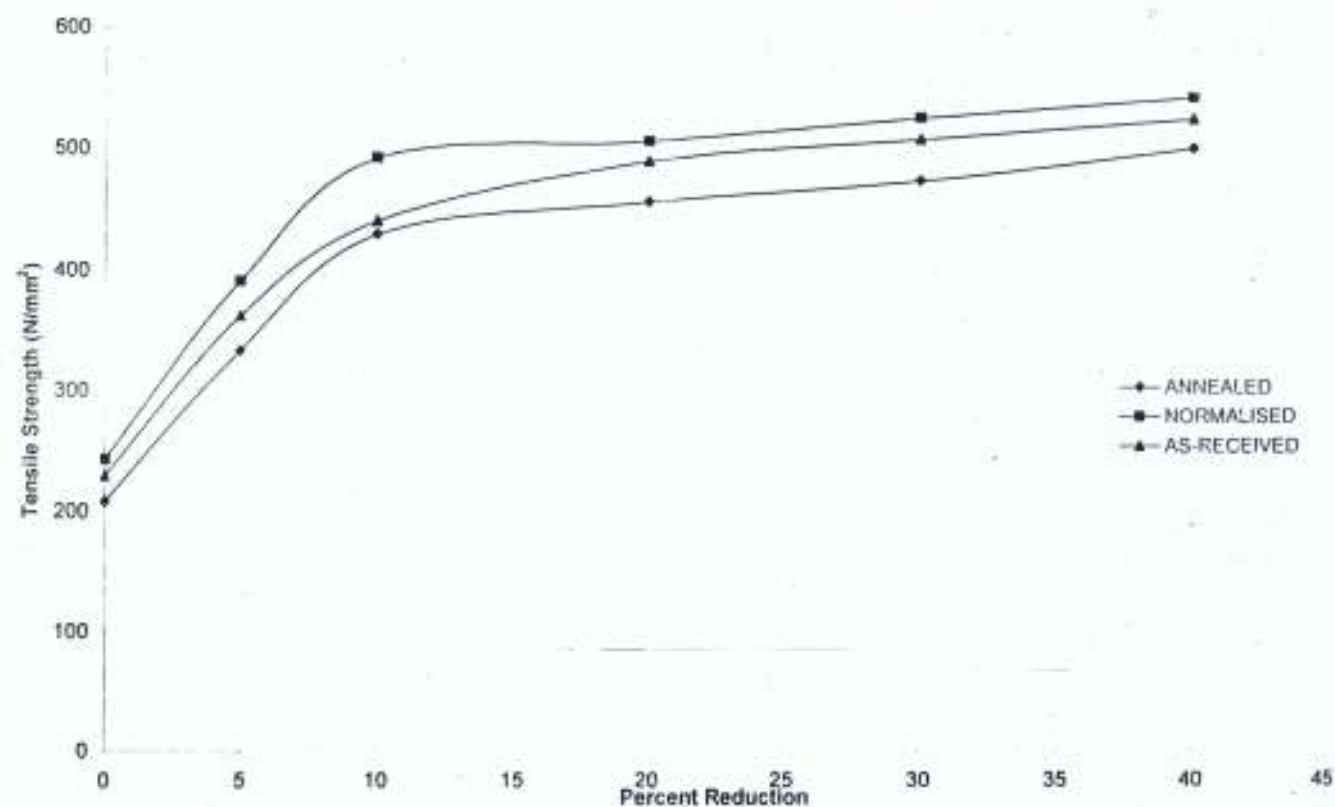


Fig:4.3(b): Tensile Strength of Welded (Weld Metal) Annealed, Normalised and As-Received Structures at Ambient Temperature (25°C)

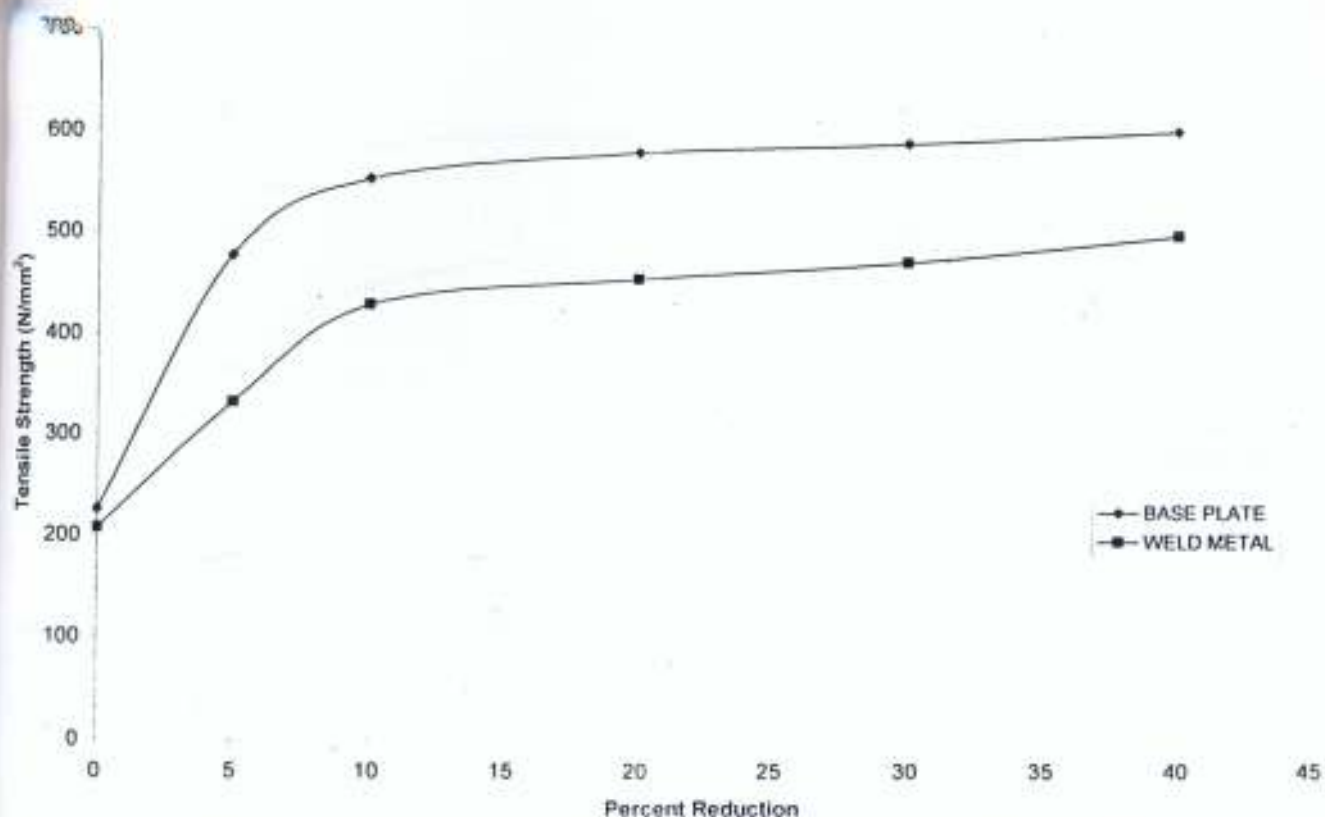


Fig. 4.3(c): Tensile Strength of Non Welded (Base Plate) And Welded (Weld Metal) Annealed Structure at Ambient Temperature (25°C)

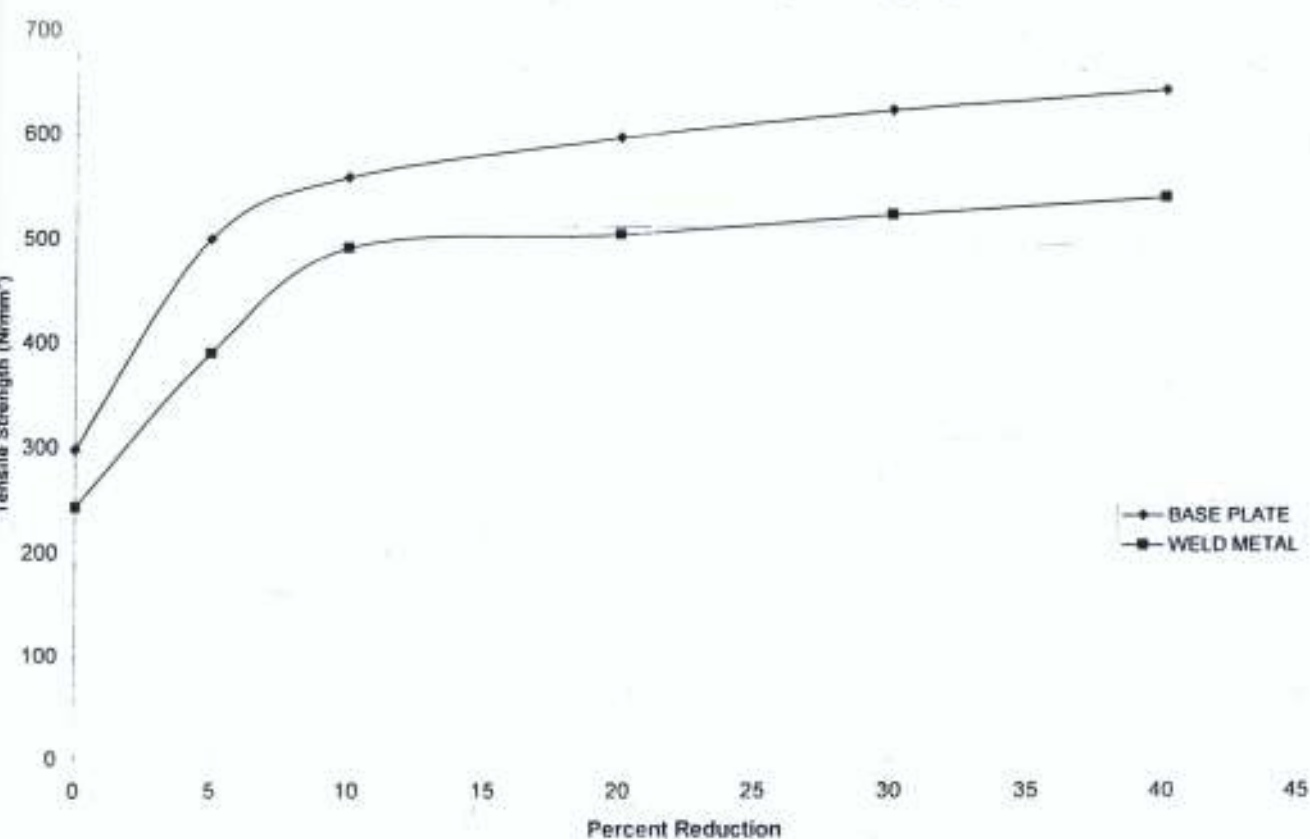


Fig:4.3(d): Tensile Strength of Non Welded (Base Plate) And Welded (Weld Metal) Normalised Structure at Ambient Temperature (25°C)

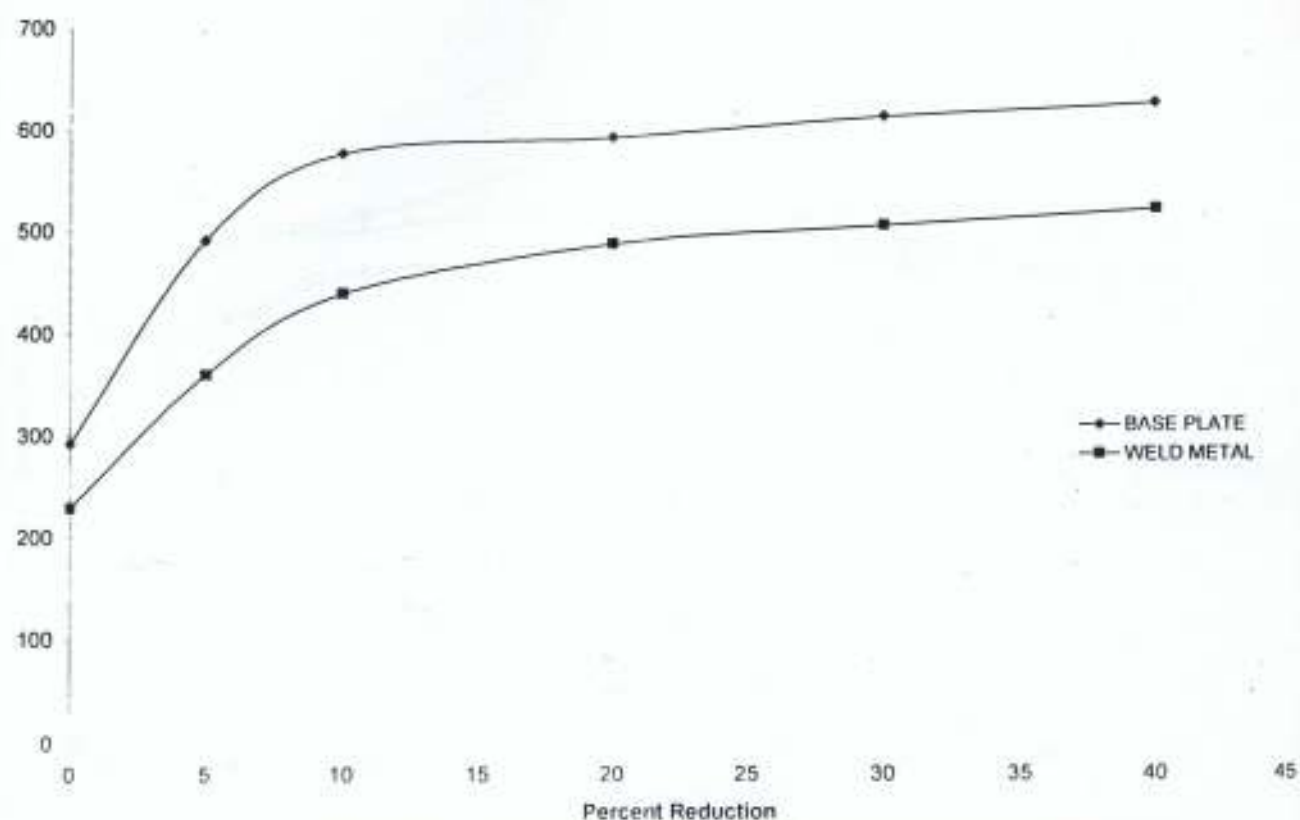


Fig.4.3(e): Tensile Strength of Non Welded (Base Plate) And Welded (Weld Metal) As-Received Structure at Ambient Temperature 25°C

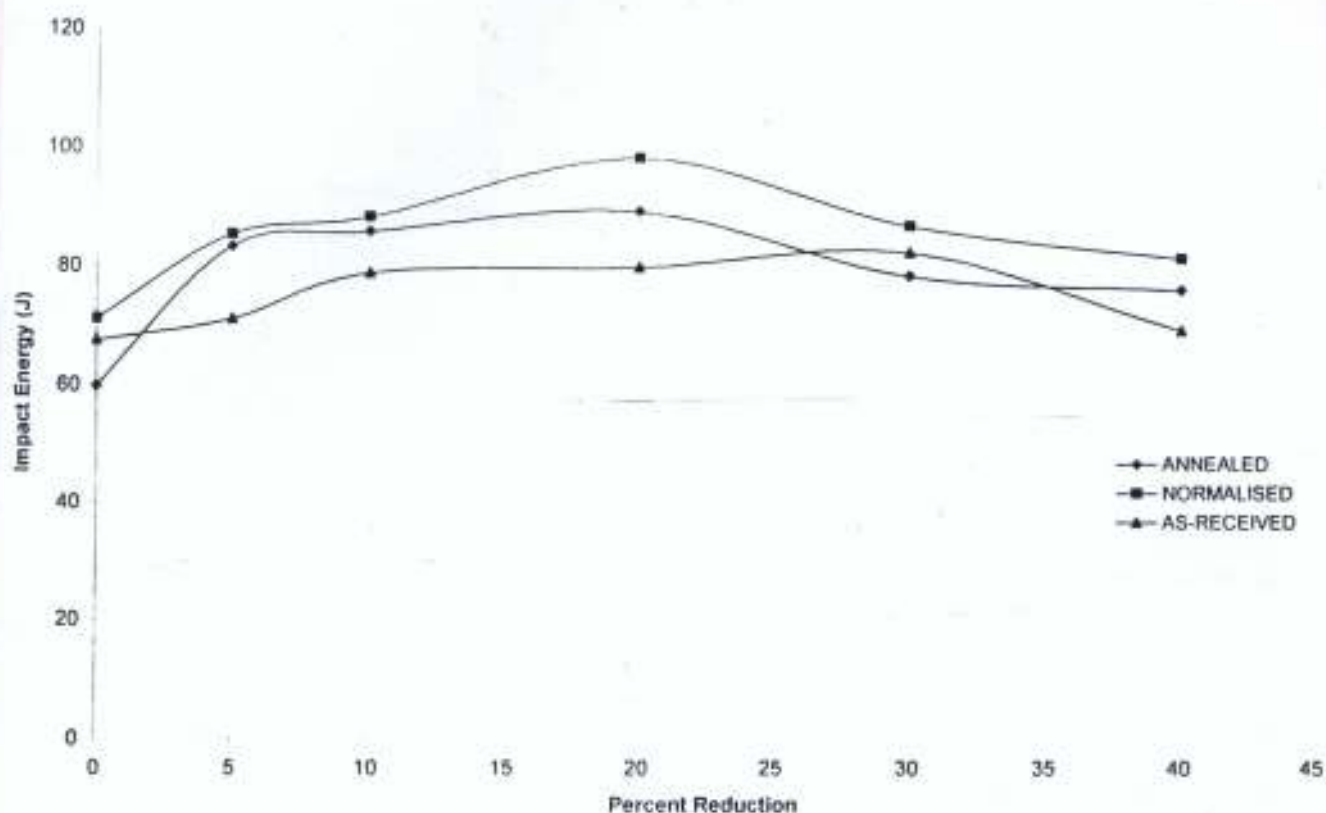


Fig. 4.4 (a): Impact Energy (Notch Toughness) of Non Welded (Base Plate) Annealed, Normalised and As-Received Structures at temperature 25°C

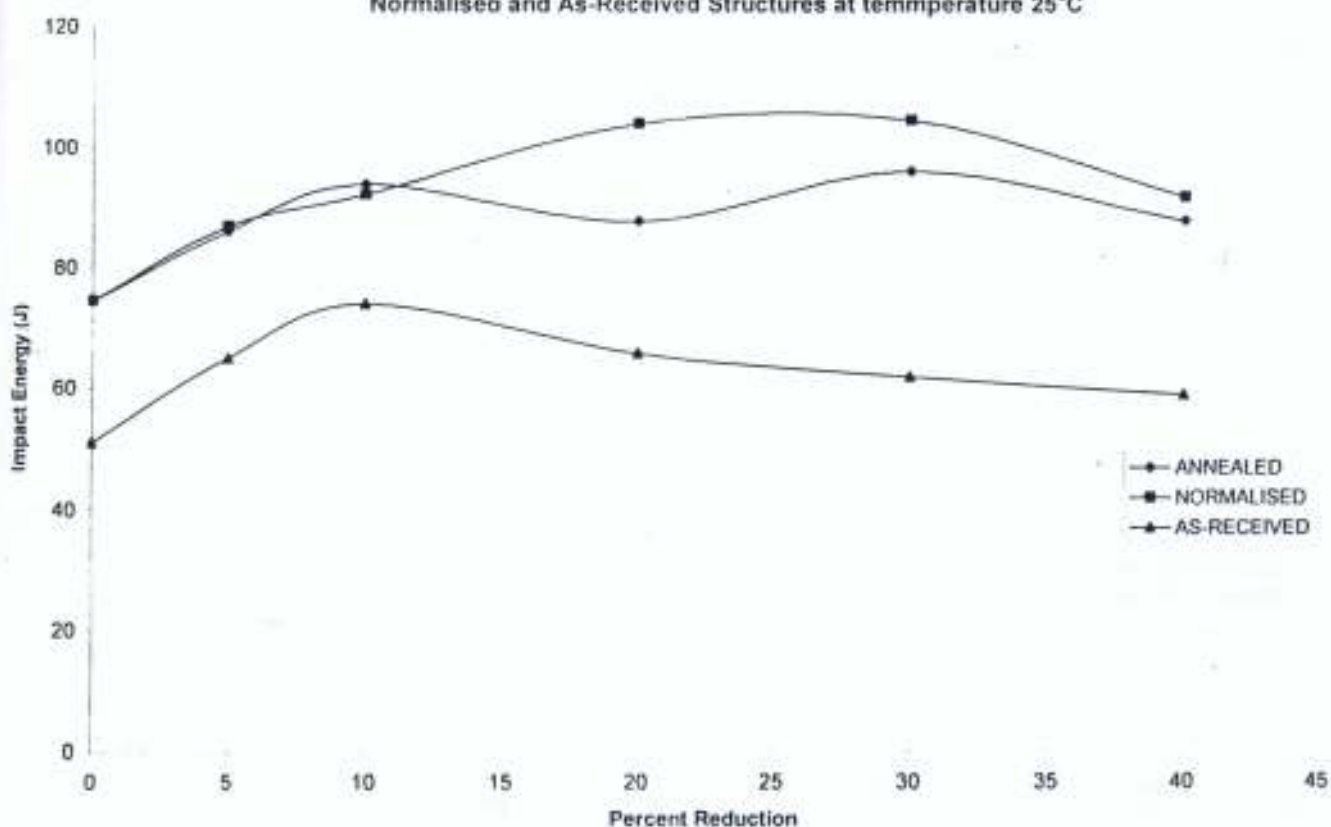


Fig. 4.4(b): Impact Energy (Notch Toughness) of Welded (Weld Metal) Annealed, Normalised and As-Received Structures at Ambient Temperature 25°C

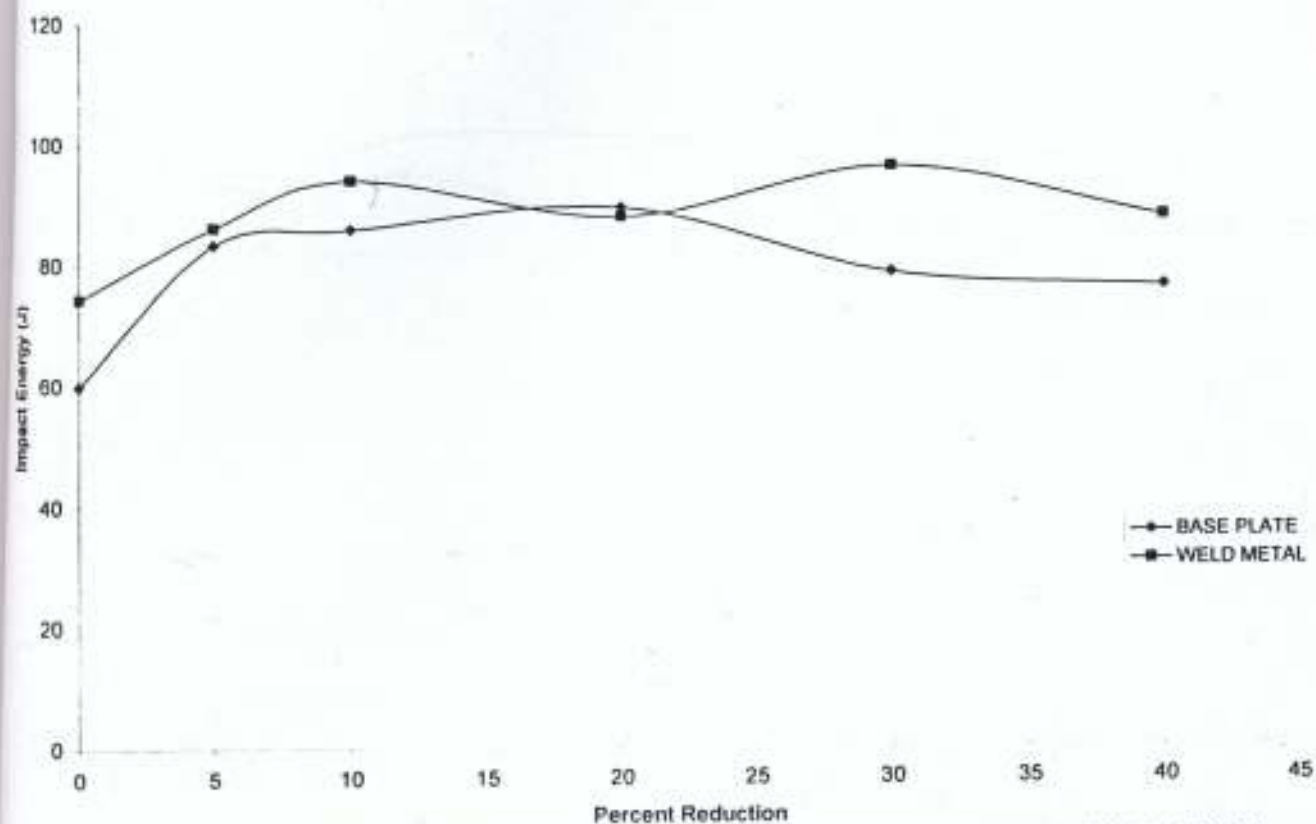


Fig. 4.4(c): Impact Energy (Notch Toughness) of Non Welded (Base Plate) And Welded (Weld Metal) Annealed Structure at Ambient Temperature 25°C

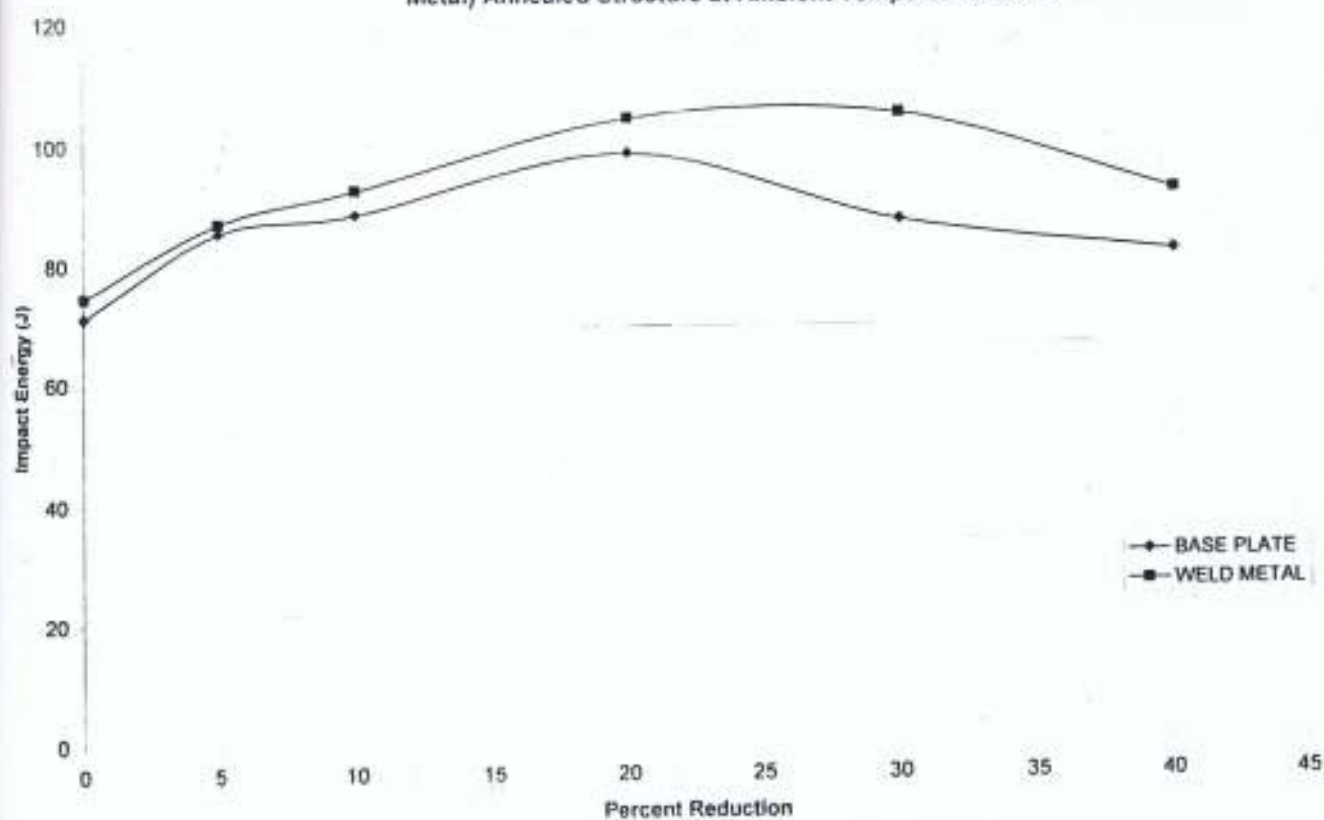


Fig. 4.4(d): Impact Energy (Notch Toughness) of Non Welded (Base Plate) And Welded (Weld Metal) Normalised Structure at Ambient Temperature 25°C

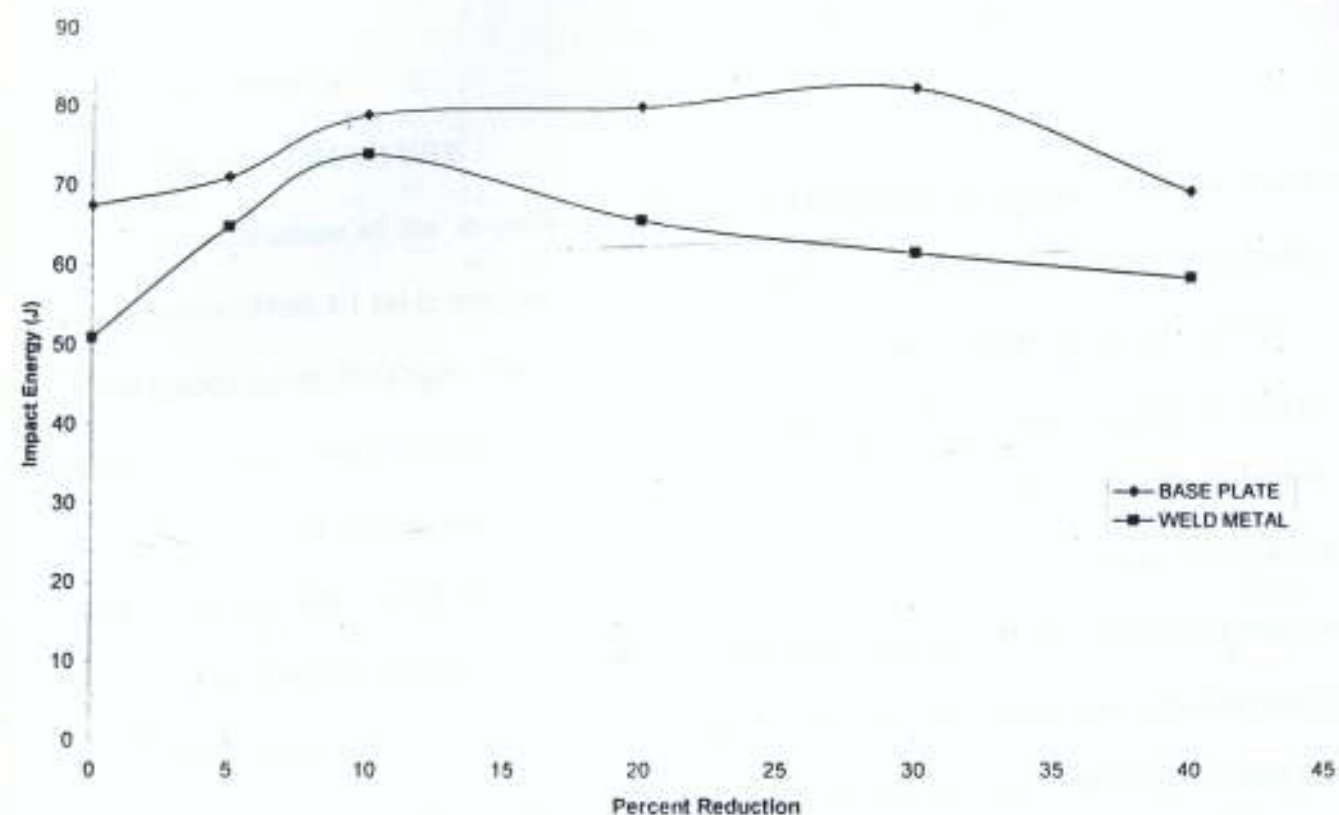


Fig. 4.4(e): Impact Energy (Notch Toughness) of Non Welded (Base Plate) And Welded (Weld Metal) As-Received Structure at Ambient Temperature 25°C

4.2 DISCUSSION

4.2.1 MICROSTRUCTURE

Microstructure of the as-received structure which consists of pearlite (dark) in ferrite (light) matrix, Plate 4.1 (a) is due the fact that pearlite and ferrite are the major microconstituents of low carbon steels, the slight elongated grains observed may have resulted from the past history of the steel plate, while the small precipitates observed might have formed during hot rolling process and did not go into solution at the austenizing temperature. Cooke et al, (2002) in their work on HSLA-100 steel, proposed precipitation of Nb (C,N) during hot rolling process, the relative more fraction volume of ferrite in annealed structure Plate 4.1 (b) could have resulted from phase transformation that occurred during heat-treatment, while the inhomogeneous dispersion of pearlite could be due to the size of austenite grains from which they transformed and the rate of cooling during the critical range[Higgin 1993].

Microstructure of the normalized structure which revealed relatively more fraction volume of pearlite phase in ferrite matrix and finer grains as compared to annealed and as-received structures Plate 4.1 (c) is due to air-cooling.

Optical microstructures of the annealed structure deformed by cold rolling to various percent (5%, 10%, 20%, 30% and 40%) reductions as presented in Plates 4.1 (a) through (e) revealed microstructures similar to those found in the non- deformed annealed structure, but with distorted pearlite in ferrite matrix, the few precipitates inherited from austenite during hot-rolling were not expected to grow on cold-working, but were found to segregate at the grain boundaries. According to Black et al, (1996) commercial metal are never pure, and due to minor segregation of the ingot casting stage, a proportion of the impurities are deposited along the grain boundaries. The steel normalized at 920°C in Plates 4.3 (a) through (e) continued to exhibit relatively greater distorted pearlite in ferrite matrix for equal percent reduction as compared to the annealed

condition, the difference is markedly seen at much higher percent reduction Plates 4.2 and 4.3 (a) through (e) compared , this is as a result of relative more fraction volume of pearlite in normalised structure. Honeycombe, (1998) showed that linear relationship exists between work hardening and pearlite content and that pearlite work hardens more than ferrite, transformation of austenite to pearlite and ferrite phases with more fraction volume of pearlite phase is due to the relative fast cooling rate associated with normalising. According to Cizek et al, (2002) the fast cooling rate takes advantage of increased steel hardenability, aims to suppress the formation of polygonal ferrite (PF) and, instead encourages non equilibrium, non equiaxed ferrite macrostructure. The latter transformation products contribute to increasing strength, through both small and effective grain sizes and increased dislocation densities, while maintaining a reasonable level of toughness.

From microstructure of the as-received structure deformed at 5, 10, 20, 30 and 40%, Plates 4.4 (a) through (e), more elongation than the one found in annealed structure is observed at corresponding percent reduction, this is due to the existence of more fraction volume of pearlite in ferrite matrix and from the work of Honeycomb, (1998) it has been shown that for equal percent reduction, pearlite work hardens more than ferrite.

Microstructures of the welded annealed structure after cold rolling at 5, 10, 20, 30 and 40 percent reduction have been provided in Plates 4.5 (a) through (e), pearlite phase in ferrite matrix is revealed. Unlike the deformed non welded annealed structure, the microstructures revealed instead of fibrous structure, new equiaxed grains with grain size decreasing as percent reduction increases. Lakhtin, (1979) had observed that when a strained-hardened metal is heated, the old grains are not restored instead entirely new equiaxed grains are formed. According to Leslie et al, (2001) recrystallized grain decreases in size with increasing cold deformation because of an increasing density of recrystallisation nuclei. The observed precipitates in the welded structure

was relatively less. Xuel et al, (1997) had also observed the formation of precipitates in their work and reported that the microstructures of HSLA steels in the weld zone are in addition, characterized by very fine carbonitride precipitates (only visible by TEM) which play a crucial role in strengthening.

Microstructures of welded normalized structure deformed at 5, 10, 20, 30 and 40 percent reduction have been provided in plates 4. 6 (a) – (e), the difference between normalized and annealed structures is seen in fraction volume and grain size of the phases difference in fraction volume of microconstituents is due to difference in the initial thermal treatments imposed on the steel while the size depends upon the degree of previous deformation, the chemical composition, the peak temperature attained and the size of the initial grain Higgin, (1993) Precipitates were observed but were relatively fewer in amounts.

Microstructures of welded as-received structures at 5%, 10%, 20%, 30% and 40% deformation is as provided in Plates 4.7(a) – (e), as in annealed and normalised structures it also revealed equiaxed grains which are basically pearlite phase in ferrite matrix, the smaller grains as compared to the corresponding annealed structure and bigger grains as compared to the corresponding normalised structures is as a result of the initial thermal treatment imposed on the steel plate. According to Sunghak et al, (1993) the resulting grain size in the weld zone would depend on the initial grain size of the base plate.

4.2.2 MECHANICAL PROPERTIES

The observed increase in hardness of the non welded (base plate) and (welded weld) metal structures with increasing percent reduction agreed with Kalpakjian, (1995) assertion that, properties of metal that are dependent on the lattice structure are affected by plastic deformation. The relative high Rockwell hardness (HRB) of the non welded part as compared to the welded parts of the various structures is due to the relative fine grains of the non welded structures .

According to Sidney, (1992) the amount of strain hardening introduced by a given amount of elongation increases as the grain size decreases, it therefore implies that hardness are significantly influenced by grain size. Gour, (1986) had showed that the non welded (base plate) during welding experience a range of temperature invariably from fusion to melting, it experience a cooling rate of the same order of magnitude as quenching, softening and recrystallisation, thus, weld zone, which cooled at different rates and includes different regions of microstructure, combines with thermal stress and strain caused by welding will yield a product different in properties from that of the base plate.

Due to growth effect of heat on grains of the welded part, hardness of the non welded part of the various structures was observed to be relatively higher than that of the corresponding welded part, and from the work of Spannos et al, (1993) on the investigation of microstructural changes in a HSLA-100 steel, micro-hardness measurement showed that the highest region of hardness occurred in region with fine grain (FG) and the lowest in region with coarse grain (CG). According to Lakhtin, (1979) hardening of metal in the process of cold deformation is due to increase in the number of defects (dislocations, vacancies, and interstitial atoms) in crystal structure, and as percent reduction increases properties increasing the resistance of metal to deformation are raised, and on the other hand, the capacity for plastic deformation is reduced.

Reis et al, (2000) investigated in their work the effect of an increasing strain on the sub structural features of low – carbon steel by comparing a 95% cold rolled steel with a steel that was cold – rolled to 70%, and observed deformed bands on the microstructure of the 70% cold-rolled sample, these bands were found to contain much less high angle boundaries ($\theta > 20^\circ$) than the ones obtained after 95% rolling strain, and concluded that a change of strain from $\xi = 1.2$ (70% rolling reduction) to $\xi = 3.0$ (95% rolling reduction) has indeed produced a significant increase of high-angle-grain-boundaries in the structure. Although hardness of non welded and welded parts of annealed structure was observed to increase with increasing percent reduction, it was however noticed that the increase rises more rapidly at the early stage of deformation and thereafter continued to increase in decreasing order, this observation could be due to the effect of annealing time on grain size which according to Black et al, (1992) the longer the annealing time the coarser the grains, and the more homogenous is the microstructure. Strong effect of annealing time on the final ferrite grains revealed that ferrite grain refinement was better when sample were submitted to annealing during 1 hour than during lesser time because a great refinement with a more uniform structure was reached [Brzussek et al, 2003]. Therefore the relatively low hardness

recorded in the non welded and welded parts of the annealed structure as compared to the corresponding parts of the normalised structure at all percent reduction is attributable to the relative coarse grain of the annealed structure. Grain size played significant role in the observed high hardness of the non welded and welded parts of the normalised structure, this is because normalising is often characterized by fine grains and well dispersed ferrite-carbide aggregates. According to Leslie et al, (2003) size of recrystallised grains decreases with increasing cold reduction because of increasing density of recrystallisation nuclei with larger strain. OLevsky et al, (2001) in their work on the constitutive respond of welded HSLA 100 steel showed by the contour of Rockwell hardness along sectional direction of a welded plate that hardness outside the weld zone is higher than that inside. Within the weld zone, it is approximately HRC 19 and HRC 25 outside the weld zone, the point close to the boundary, between the weld metal and the HAZ, showed higher hardness than the weld center area. The higher hardness values of non welded and welded parts of the as-received structure as compared to the corresponding parts of the annealed structure is due to grain size effect, while the initial rapid rise in hardness of the welded part of the annealed structure at low degree of deformation, is attributed to metallurgical changes, controlled by cooling rate, the noticed irregularity in hardness of the welded part of the as-received structure is as a result of inhomogeneous dispersed carbide within the matrix. Generally, increase in hardness resulting from increasing deformation is according to Lahktin, (1979) based on the concept that grain boundaries act as barriers to dislocation motion. The relative increase in hardness of non welded part and welded part of normalized structure as compared to non welded and welded parts of annealed and as-received structures may be attributed to metallurgical factors which according to Kalpagian, (1995) includes grain orientation other reasons include more fraction volume of pearlite in ferrite matrix, well dispersed ferrite-carbide aggregates and grain size effect.

Sidney, (1982) had observed that for equal amount of cold working more strain hardening is introduced into initially fine-grained steel than into initially coarse-grain steel. Therefore, non welded and welded parts of the as-received structure with a relatively smaller grains exhibited superior hardness over the corresponding non welded and welded parts of annealed structure, especially at high percent reduction.

The observed increase in tensile (TS) and yield strength (YS) properties of normalized, annealed and as-received structures with increasing percent reduction as in hardness is largely due to grain size effect. Rajan et al ,(1988) have showed that grain size significantly influences

the mechanical properties of metals. At room temperature, large grain is generally associated with low strength. Lakhtin, (1979) asserts that high strength is associated with smaller grain because they have a larger grain – boundary surface area per unit volume of metal.

Lancaster, (1993) had observed the importance of heat input rate in determining the growth of recrystallized grains in fusion welding as this governs heating rates, cooling rates and weld pool size. Low input rate process which gives relatively high cooling rate process generate fine grains and less embrittlement compared to high input rate process with longer thermal cycle and coarser structure. The superior tensile and yield strength exhibited by the non welded and welded parts of normalized structure as compared to the corresponding parts of annealed and as – received structures is attributed to normalizing effect. Lakhtin, (1979) showed that the more cooling in air used in normalizing causes the austenite to decompose at lower temperatures, the superior strength of the non welded parts of the structures over the corresponding welded parts of the structures according to Que et al, (2002) on constitutive response of welded HSLA100 is due to relative fewer defects in the non welded part as compared to welded part. Kalpagian, (1995) had observed that the weld is a cast structure which is often associated with impurities.

Although there occurred a continuous increase in yield and tensile strength in non welded and welded parts of the as-received structure over the entire range of percent reduction Fig. 4.4(c) its values were relatively lower than that of the corresponding parts of normalized structure and higher than that of the annealed structure, this observation could be due to the heat treatment the steel plate was given. At increasing percent reduction, yield and tensile strength were improved considerably for the non welded and welded parts of the structures, this is to be expected because on welding equiaxed grains are formed in place of fibrous structure and the size of these grains depend on the degree of deformation given to the base plate prior deformation. Fujioka, (1979) observed that the greater the degree of previous deformation, the smaller the recrystallized grains. The observed initial relative rapid rise in yield as compared to tensile for the non welded and welded parts of the annealed, normalised and as-received structures agreed with Black's, (1996) observation; that an initial rapid rise in yield occurs such that, as amount of plastic deformation is increased, the gap between yield and tensile decreases

The superior impact energy (notch toughness) exhibited by welded and non welded parts of the annealed and normalized structures over the corresponding parts of the as-received structure at all percent reduction is due to improved grain size brought about by heat treating Fig. 4.4(a) and Fig. 4.4(b) compared, and the observed superior impact energy (notch toughness) of

the non welded part over the welded part of the as received structure Fig.4.10 (c) is as a result of initial small grain size of the as-received structure prior to heat treatment. Rajan, (1989) observed that increase in grain size raises impact transition temperature, and thus makes the steel more prone to failure by brittle fracture. Non welded part of normalized structure Fig.4.4 (b) exhibited less superior impact energy (notch toughness) over corresponding welded part at all percent reduction, this in addition to grain size effect could be as a result of sample dimension and type of electrode employed. Mujahid et al, (2003) attributed fall in impact energy (notch toughness) to both negative effect of matrix strengthening and impurities segregation at grain boundaries. The unsteady rise in impact energy (notch toughness) over the entire percent reduction of welded parts of the structures is due to inhomogeneous dispersion of pearlite (dark) in ferrite (light) matrix. Higgin, (1993) observed that phase distribution, alloying elements and impurities all have effects on impact (notch toughness) energy. The superior impact energy (notch toughness) of welded part of the annealed structure over corresponding part of the normalized structure especially at higher percent reduction Fig. 4.4(b)) could have resulted from more fraction volume of pearlite in ferrite matrix as compared to the welded part of the annealed structure. Honeycombe, (1998) showed that pearlitic structures are characterized by many crack nuclei at the pearlitic interfaces.

The low impact energy (notch toughness) of welded part of as-received structure compared to welded parts of annealed and normalized structures is due to large frequency of void or microcracks initiation which are removed upon annealing and normalizing heat treatment processes.

CHAPTER FIVE

5.0 CONCLUSION AND RECOMMENDATIONS

5.1 CONCLUSION

The welded part of the normalized structure exhibited best combination of hardness, yield and tensile properties over the corresponding welded parts of the annealed and as-received structures, these properties were observed to increase with increasing percent reduction. However, its highest impact toughness was obtained at less than 20 percent reduction, at higher percent reductions (30%, 40% & presumably above) welded part of the annealed structure was observed to exhibit improved impact energy (notch toughness) over corresponding welded part of the normalized structure. The observed overall low energy (notch toughness) absorbed on impact by the welded part of the normalized structure at increasing percent reduction is attributable to more fraction volume of pearlite in ferrite matrix, pearlitic structures are generally characterized by many crack nuclei at pearlitic interfaces which upon impact restricts plastic deformation in the vicinity of the cracks.

5.2 RECOMMENDATIONS

Based on the experimental results, it is seen that under slow condition of loading the best combination of mechanical properties was obtained with the normalised structure and under fast condition of loading (impact) with annealed structure. Therefore, it is recommended that the type of heat treatment these steels are given should be based on the operating condition under which they would be used. Also, similar studies should be carried out on low carbon steels of various chemical compositions. Apart from arc welding other available welding processes should be employed as well. Since the mechanical properties of these steels are determined by their

microstructural characteristics, it is suggested that the use of transmission electron microscope (TEM) should be encouraged in subsequent studies.

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APPENDIX I**TABLE 4.1 (a-c): HARDNESS OF (NON WELDED) AND WELDED (WELD METAL) ANNEALED, NORMALISED AND AS RECEIVED STRUCTURES**

PERCENT REDUCTION	ANNEALED (BASE PLATE) N/mm ²	ANNEALED (WELD METAL) N/mm ²
0	49.65	46.19
5	63.36	59.25
10	75.69	65.94
20	77.70	68.44
30	79.92	69.22
40	75.01	70.37

PERCENT REDUCTION	NORMALISED (BASE PLATE) Nmm ²	NORMALISED (WELD METAL) Nmm ²
0	54.63	51.26
5	67.31	62.39
10	79.92	72.73
20	82.91	74.86
30	83.09	75.71
40	83.99	76.49

PERCENT REDUCTION	AS RECEIVED (BASE PLATE) N/mm ²	AS RECEIVED (WELD METAL) N/mm ²
0	50.73	47.48
5	63.34	62.19
10	76.61	67.89
20	77.38	68.54
30	77.84	70.30
40	79.99	73.36

APPENDIX II

TABLE 4.2 (a-c): YIELD STRENGTH OF NON WELDED (BASE PLATE) AND WELDED (WELD METAL) ANNEALED, NORMALISED AND AS- RECEIVED STRUCTURES

PERCENT REDUCTION	ANNEALED (BASE PLATE)Nmm ²	ANNEALED (WELD METAL) N/mm ²
0	158.01	96.27
5	339.22	217.00
10	426.11	279.44
20	461.31	309.59
30	483.04	332.80
40	564.10	350.15

PERCENT REDUCTION	NORMALISED (BASE PLATE)Nmm ²	NORMALISED (WELD METAL) N/mm ²
0	163.92	135.02
5	348.70	303.31
10	435.89	370.35
20	477.59	387.83
30	520.59	417.31
40	549.08	437.91

PERCENT REDUCTION	AS- RECEIVED (BASE PLATE) Nmm ²	AS- RECEIVED (WELD METAL) N/mm ²
0	167.15	132.70
5	376.28	280.08
10	468.25	326.82
20	491.54	351.06
30	514.28	394.25
40	535.26	427.83

APPENDIX III

TABLE 4.3 (a-c): TENSILE STRENGTH OF NON WELDED (BASE PLATE) AND WELDED (WELD METAL) ANNEALED, NORMALISED AND AS- RECEIVED

PERCENT REDUCTION	ANNEALED (BASE PLATE)N/mm ²	ANNEALED (WELD METAL) N/mm ²
0	225.61	207.42
5	477.02	332.42
10	552.91	429.01
20	579.13	455.12
30	588.95	472.43
40	601.77	499.22

PERCENT REDUCTION	NORMALISED (BASE PLATE)N/mm ²	NORMALISED (WELD METAL) N/mm ²
0	297.18	242.06
5	499.68	390.19
10	559.13	491.55
20	597.21	505.27
30	623.86	524.38
40	643.64	541.37

PERCENT REDUCTION	AS- RECEIVED (BASE PLATE) N/mm ²	AS- RECEIVED (WELD METAL) N/mm ²
0	292.51	228.80
5	491.31	360.82
10	577.20	440.12
20	593.16	488.88
30	613.89	506.80
40	627.62	523.81



APPENDIX IV

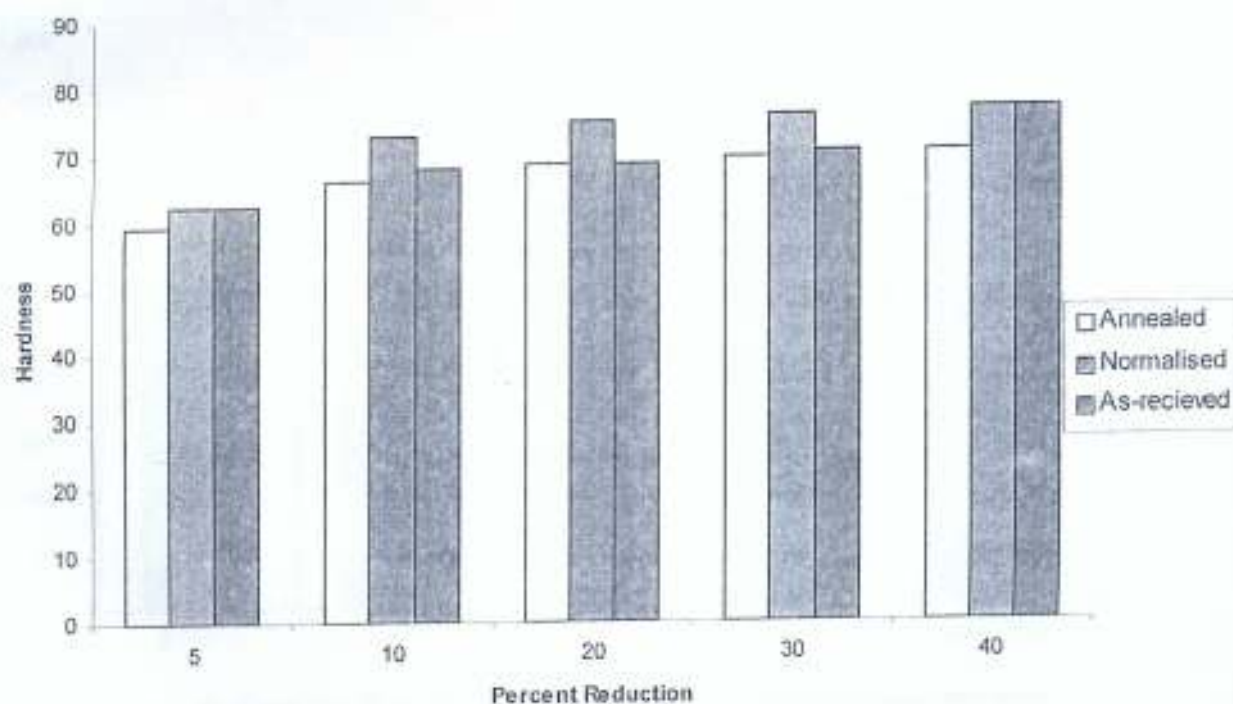
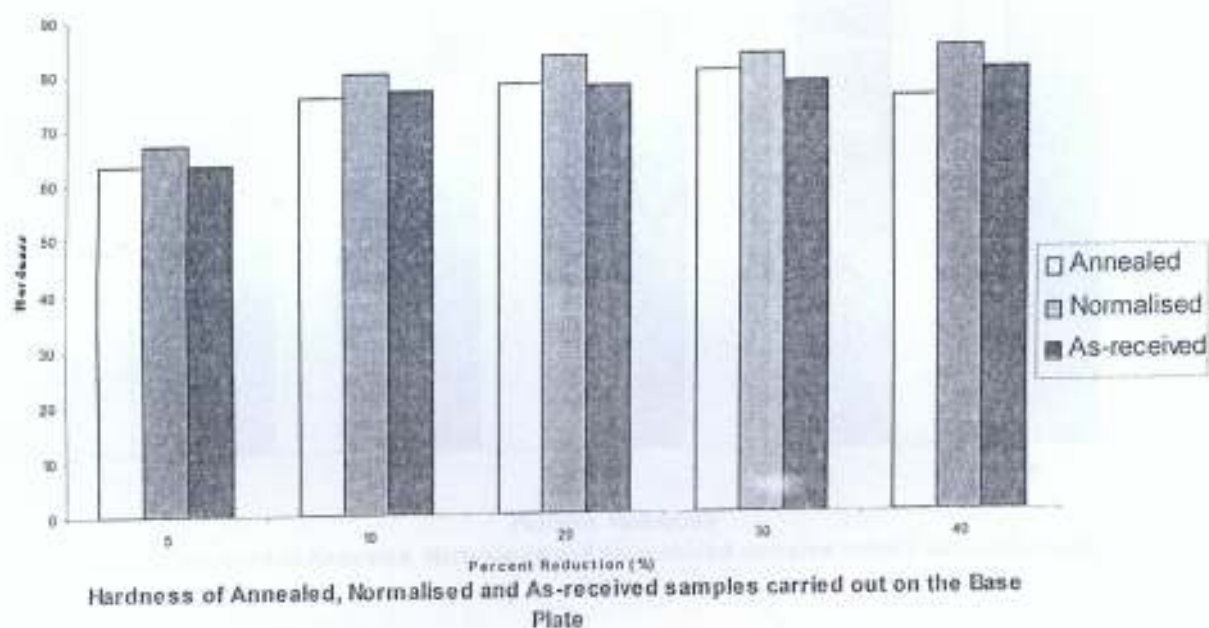
TABLE 4.4 (a-c): IMPACT (NOTCH TOUGHNESS) ENERGY OF NON WELDED 9BASE PLATE) AND WELDED (WELD METAL) ANNEALED, NORMALISED AND AS-RECEIVED STRUCTURES

PERCENT REDUCTION	ANNEALED (BASE PLATE) J	ANNEALED (WELD METAL) J
0	59.76	74.33
5	83.29	86.09
10	85.82	93.93
20	89.36	87.84
30	78.64	96.09
40	76.36	87.85

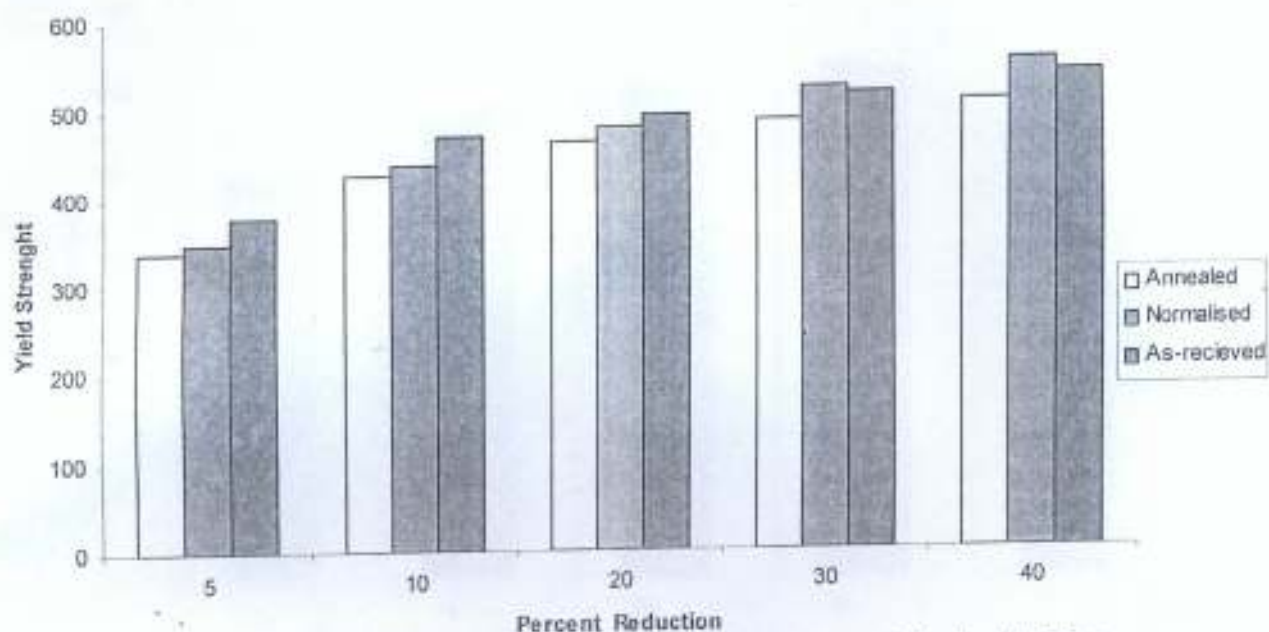
PERCENT REDUCTION	NORMALISED (BASE PLATE) J	NORMALISED (WELD METAL) J
0	71.02	74.45
5	85.32	86.92
10	88.31	92.32
20	98.34	104.11
30	87.03	104.67
40	81.61	91.86

PERCENT REDUCTION	AS- RECEIVED (BASE PLATE) J	AS- RECEIVED (WELD METAL) J
0	67.45	50.89
5	71.10	64.90
10	78.90	73.96
20	80.03	65.82
30	82.56	61.88
40	69.58	58.87

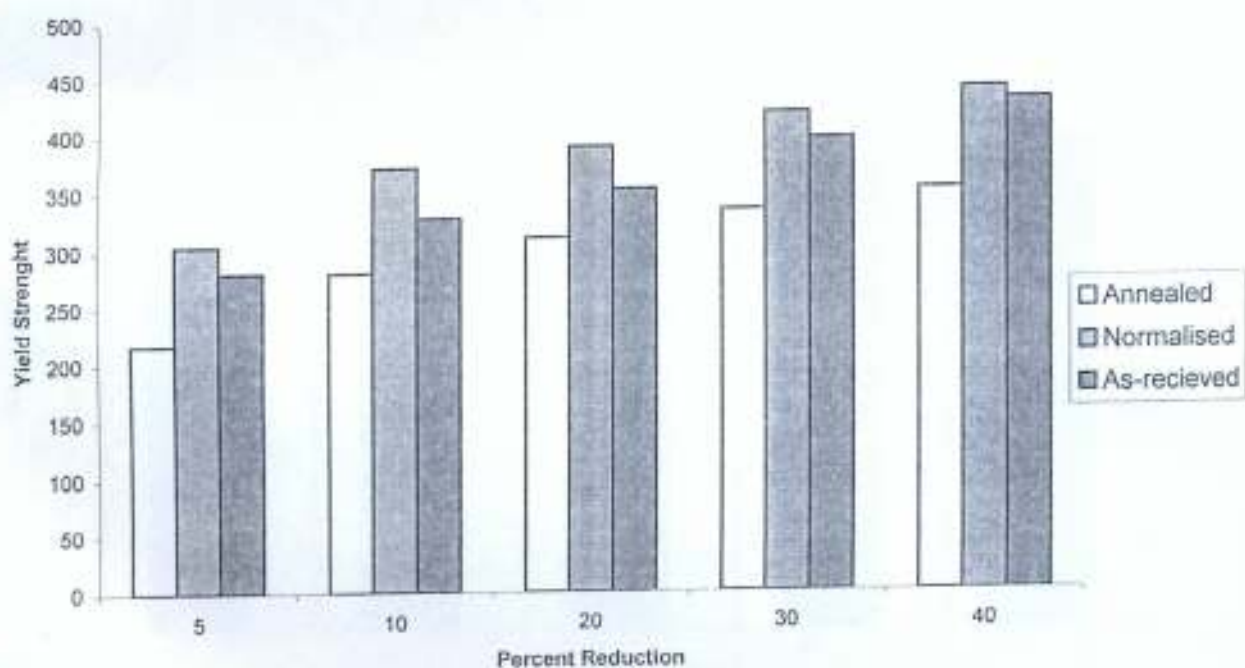
APPENDIX V



APPENDIX VI

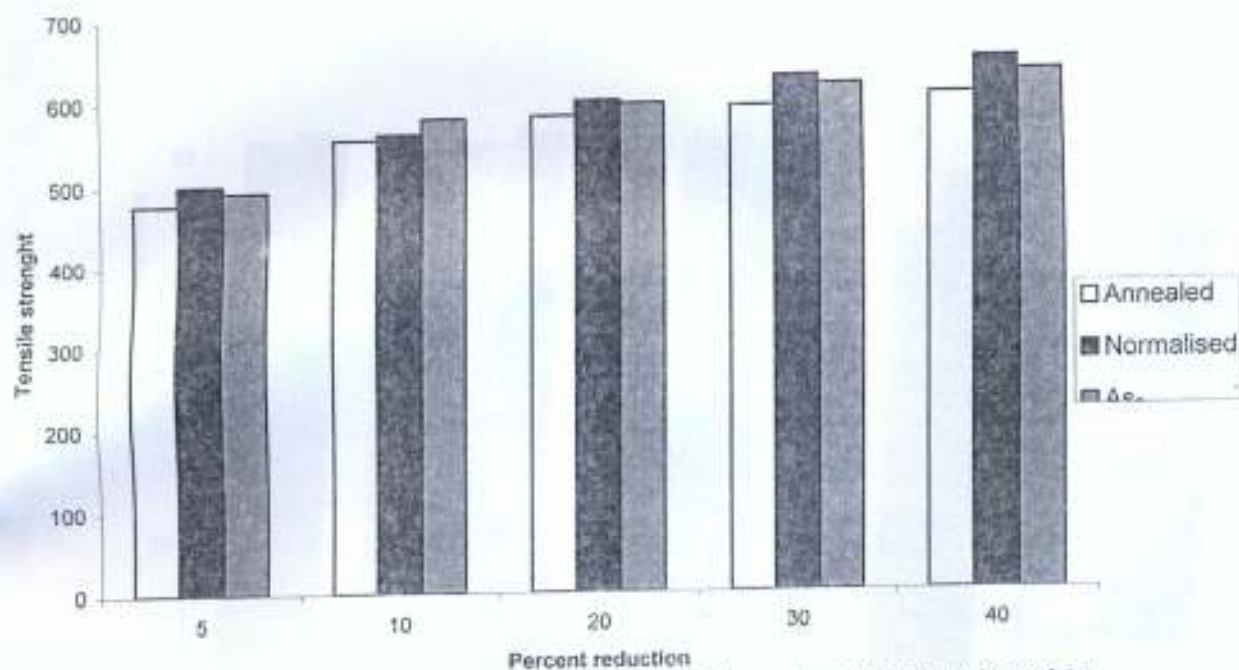


Yield Strength of Annealed, Normalised and As-received samples carried out on the Base Plate

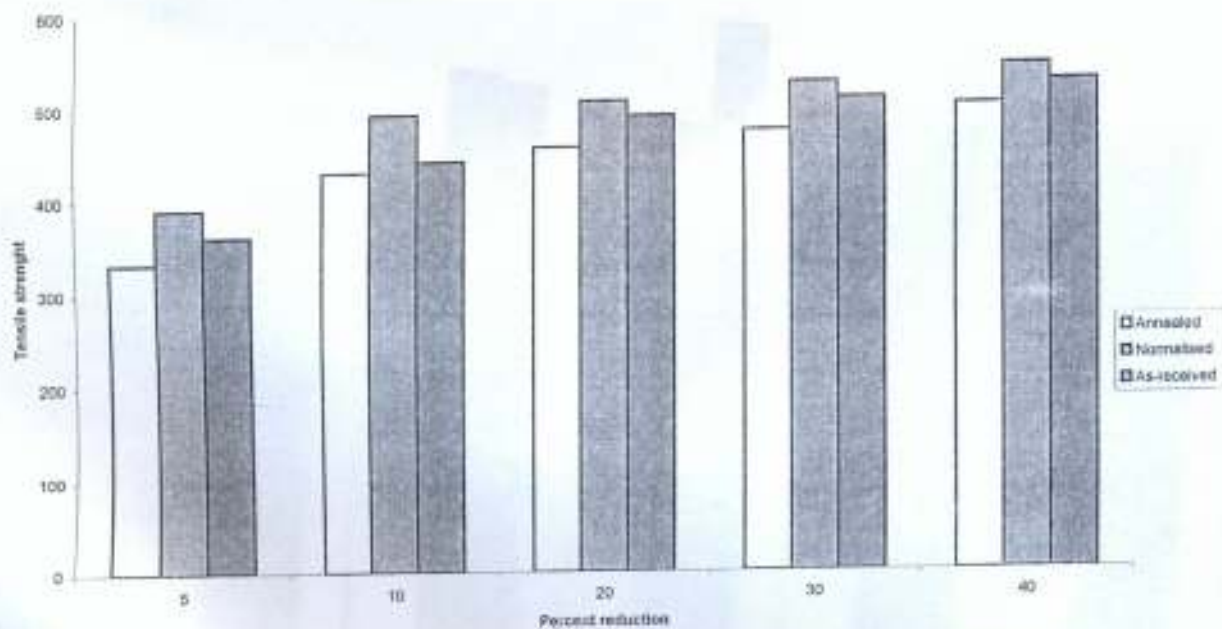


Yield Strength of Annealed, Normalised and As-received samples carried out on the Weld Metal

APPENDIX VII

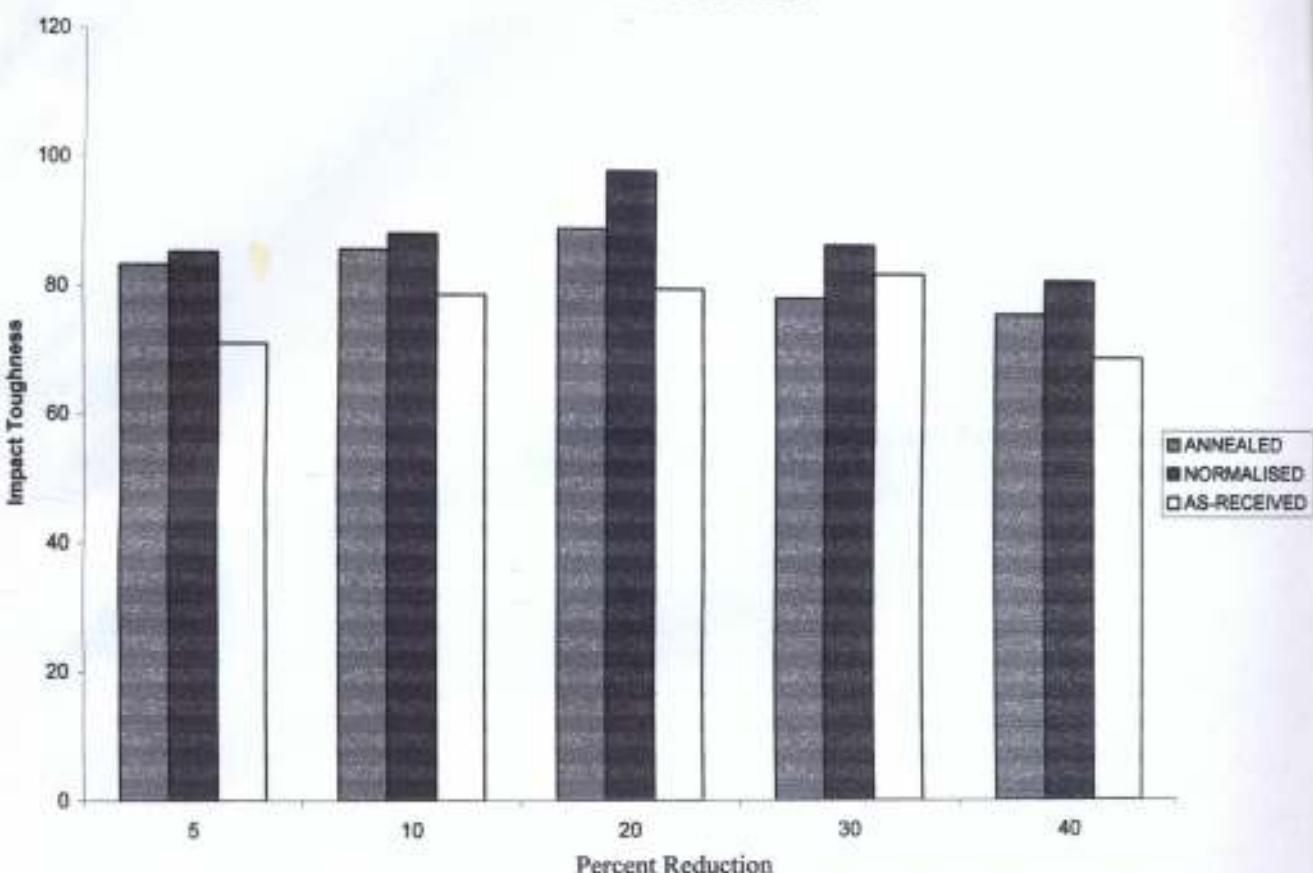


Tensile Strength of Annealed, Normalised and As-received samples carried out on Base plate

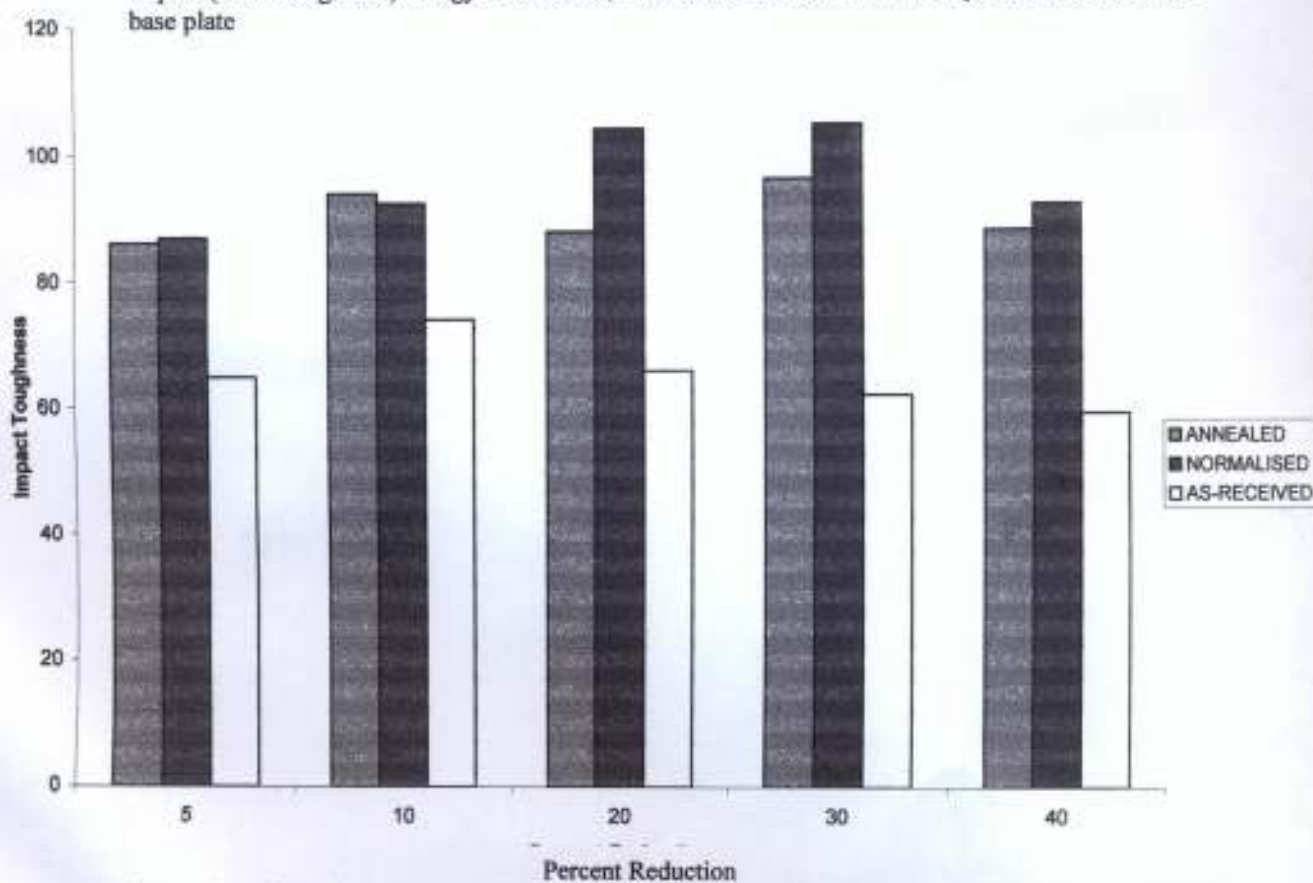


Tensile strength of Annealed, Normalised and As-received samples carried out on Weld Metal

APPENDIX VIII



Impact (notch toughness) energy of annealed, normalized and as-received samples carried out on the base plate



Impact (notch toughness) energy of annealed, normalized and as-received samples carried out on the weld metal