"THE STUDY OF HARDENING CHARACTERISTICS OF HIGH CARBON STEELS AND DUCTILE CAST IRON"

BY

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A thesis submitted to the Postgraduate school, Ahmadu Bello University, in partial fulfillment of the reguirements for the degree of Master of Science(Metallurgical Engineering).

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CERTIFICATION

This thesis entitled
"The Study of Hardening Characteristics of High Carbon Steels and Ductile Cast Iron"

By

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Meets the regulations governing the award of the degree MASTER OF SCIENCE(Metallurgical Engineering) of Ahmadu Bello University, and is approved for its contribution to knowledge and literacy presentation.

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DECLARATION

I hereby declare that this thesis title "The Study of High Carbon Steels and Ductile cast iron" contains the report of my research work and has not been presented/accepted in any previous application for higher degree. *Information from other sources has been acknowledged by means of reference.*

DEDICATION

This work is dedicated with love to my wife, Rose, my children, Doimen and Gyeyock and to the memory of my beloved late mother, Mrs Hauwa Dauda.

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I am very much grateful to Dr J.O.T. Adewara, my supervisor for his immense contributions, guidiance and useful criticism through out the duration of this work. The inputs of my second supervisor, Dr O. B. Oloche is immeasurable and well appreciated.

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Above all I am most grateful to God Almighty the father of our Lord Jesus Christ who gave me the strength to hold on till the end of this work.

ABSTRACT

The hardening characteristics of four grades of high-carbon steel materials and ductile iron material were studied. Three of the steel materials were plain high carbon steels with compositions of 0.95%C, 1.05XC and 1.20XC. the fourth steel material was a low-alloy, with 1.205%C and 1.20%W.

Standardized specimens were cut from these materials and were first given a normalizing heat treatment before they were hardened. Hardening was done by heating to an appropriate temperature followed by quenching in water or oil. Hardness measurements were made on these specimens in the as quenched condition or after further conditioning treatments. These included:

(a) Tempering between 150 and 500^{o}C immediately after quenching,

(b) Ageing,

(c) Cold-treatment and tempering.

Other tests carried out on these materials include Jominy end-quench tests, to determine their hardenabilities, and notched impact tests to determine their toughness both in the as-quenched condition and after tempering at temperatures between 150 and 500^{o}C.

All the steel materials responded to hardening by water quenching, but the ductile iron material did not. Only the low-alloy high carbon steel responded to hardening by oil-quenching. It was also found that optimum hardening temperature for these steels was between 800 and 850^{o}C. Tempering characteristics of the steel materials showed that they all soften slightly on heating to 250^{o}C presumably due to stress relief. Beyond this temperature softening was rapid for the plain carbon steels but was arrested in the case

of the low-alloy steel for up to 500^{o}C covered In this work.

Cold treatment after quenching caused remarkable increase in the hardness of the high carbon steel materials, both in the tempered and untempered condition. This was probably due to the transformation of the retained austenite during the cold treatment. Ageing treatment which occurred tor over two months was also found to cause the same effects.

Tempered notched impact tests produced toughness peaks at 350^{o}C In the steel specimens. This gave an optimum tempering temperature for maximum toughness over a narrow range around this temperature, for the high carbon steels which were water quenched and followed immediately by tempering. The critical diameters of the plain high-carbon steels were found to be 16mm, 14mm and 12mm for the 0.95XC, 1.05XC and 1.20XC carbon steels respectively. These confirmed the low hardenability character of plain high carbon steels. However it was 65mm for the W-based low-alloy high carbon steels.

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CHAPTER ONE

1.0 INTRODUCTION

In the early civilization, steels were used to make tools to supplement the strength of man. In this respect man had fashioned weapons to protect himself against enemies, and produced machines to process food and materials for health, comfort and pleasure of man. Later, steel became the key reason for advancements in transportation, industrialization and economic development of any modern society, Roberts et al (1961),

Today, according to Payson (1962), civilization is based on the advanced machines which depend on the properties of metallurgical engineering materials for high strength, toughness and hardness. It is noteworthy that these machines would have been impossible if materials for cutting, forming and shaping of the various component parts were not developed.

The development of modern tool steels can be traced to the eighteenth century, with the rediscovery of high temperature melting practice by Huntsman, Roberts et al (1961). However, high rate of machining could not be done before the development of the mushet's special steel containing 2% carbon(C), 7% Tungsten(W) and 2.4% Manganese(Mn). Prototype of the modern high speed steels came to use in 1903, Townsend (1933). However, it was not until after the second world war when the technology for modern high speed was clearly defined, Payson (1962).

Tool steels are classified on the basis of their applications for forming, molding, shearing, cutting and other miscellaneous operations. The tool steels for cutting and forming operations are high-carbon steels in the alloyed and unalloyed conditions. The cutting ability of these steels is improved by alloy additions.

In Nigeria the use of tool steels are common for domestic and industrial applications, and these include tools for:

a. Domestic Usage:- cutlass, knives, agricultural implements, razor blades and among others.
b. Work Shop Practice:- Wood saws, hack saw blades, shearing blades, drills, cutters, punches, files, hammers and so forth.
c. Industrial Productions:- rolls, dies for forming, extrusion, drawing, high speed cutting tools.

A good proportion of tools used in Nigeria for domestic and workshop operations are plain high carbon steels. Tools for industrial productions are all, without exception, alloyed steels operating under hash conditions of high hardness at elevated temperatures. They are much more expensive and require strict control measures for production. Virtually all these tools possessing suitable qualities are presently imported into the country. Others which are produced in the country are made from improper materials and without due attention being paid to heat treatment requirements. Mild steels are often employed in places of the normal and appropriate grade of tool steels.It is therefore important to look into the manufacture of the common

workshop tools using the ordinary plain high carbon steels and or ductile cast iron. The inclusion of the ductile cast iron comes from the fact that its structure consist of graphite nodules in a high carbon steel matrix.

CHAPTER TWO

2.0 LITERATURE REVIEW

2.1.0 General Characteristics of Steel

In general, steels can be classified into the plain and alloy steels. The plain carbon steels are basically Iron(Fe)-Carbon(C) alloys which often contain Silicon(Si), Manganese(Mn), Sulphur(S) and Phosphorus(P) which are introduced either as a direct consequence of the production process or intentionally for enhanced mechanical properties. The total additions of these elements in carbon steels are normally low. Generally composition for these steels requires maximum values of 1.50%C, 1.50% Mn, 1.50% Si, 0.03%S, and 0.03%P.

Generally, alloying elements are employed for high mechanical properties (strength, hardness and toughness), low and high temperature properties, corrosion, magnetic and electrical properties, among others. Alloy steels therefore are those which derive their enhanced properties from the presence of one or more special element(s), such as Copper(Cu), Manganese(Mn), Silicon(Si), Chromium(Cr), Nickel(Ni), Molybdenum(Mo), Tungsten(W) and Vanadium(V). Under certain circumstances, Aluminum (Al), Titanium(Ti), Niobium(Nb), Tantalum(Ta), are usually added in small quantities, Bain and Paxton (1961).

2.1.1 Classes Of Carbon Steels

The following classes of carbon steels are of great importance:

(a) Excellent formability Steels

These are extremely low carbon steels with carbon(C) less than

0.15%. They find applications among others, such as, cold-rolled sheets for car bodies, roofing sheets and soft magnets.

(b) Engineering Construction Steels

These contain higher carbon contents between 0.1 and 0.3%C and are usually used in the hot rolled or normalized condition for excellent strength, ductility and good machinability. Hence they are used as welded structures for bridges and buildings, ship plates, boiler plates, forgings among others.

(c) Machine Building Steels

These constitute steels with 0.25-0.65%C and respond strongly to strengthening by quenching and tempering. The (0.25-0.45%C) grades are commonly used as forgings which attain superior mechanical properties than when they are quenched and tempered only.

The 0.40 - 0.65%C grades have excellent hardenability and are used in the hardened and tempered condition. Hence are for relatively highly stressed components for example machines elements such as gears, shafts, axles, pinions keys and so forth.

(d) Shock Resistant steels:- are strong, tough, wear resistant and *are generally, of high elastic properties. They are used for* springs piston rings, chisels, hard tools among others, Dell (1978). The carbon content in usually between 0.5 and 0.9%C.

(e) Wear Resistant Steels:-

These are the hyper-eutectoid steels for which residual carbides are retained.it is usually employed in the hardened and tempered condition for which its wear resistance is obtained from the

presence of residual carbides and tempered martensite. They are classified into three grades:

(i) 0.8 - 1.00%C grade:- is used for high hardness and high tensile strength, areas of application include springs, high tensile wires,knives,axes, dies for all purposes, metal punches and rock drills.

(ii) 1.00 - 1.20%C grade:- is used for tools where hardness is of primary consideration, as in ball bearing, cold cutting dies, drills, wood working tools, lathe tools, milling cutters, knives among others.

(iii) 1.20 - 1.50%C grade:- is used where a keen cutting edge is necessary such as razors, hack saw blades, shearing blades, blanking dies and so forth.

Carbon steels are generally low heat steels for which the mechanical properties decrease rather rapidly at temperatures above 250°C, Dell (1978). This is especially the case with the high carbon steels in excess of 0.65%C, which have relative low softening temperatures for martensite. In addition, they are shallow hardening steels and the addition of alloying elements readily improves their properties.

2.1.2. Alloy Steels

When alloying elements are added in small quantities to steels, they are known as low alloy steels and the total amount of alloying elements is usually less then 5%, Bain and Paxton (1966). These steels find applications in areas requiring:-

(a) High strength,

(b) Improved hardenability,

(c) Low temperature resistance,

(d) Creep resistance.

Alloying addition in excess of 5% results in the formation of higher alloy steels and the properties imparted to these steels include:

(a) Ultra high strength,

(b) High hardness at low and high temperature,

(c) High temperature resistance,

(d) Corrosion resistance.

According to Pickering (1977), the principles involved in alloying of steels includes:

(a) Specific property requirements,

(b) Condition of applications of the steel,

(c) Type of alloying elements chosen,

(d) Cost and availability considerations.

Basic principles in the design of alloy steels for high strength applications depend on the composition of carbon. Different strengthening mechanisms are involved for the low carbon, medium carbon or high carbon alloy steels, Pickering (1977).

In the case of structural steels, alloy design is based on the C-Mn-Si systems. The high strength property is obtained by fine grain structure, controlled mechanical working, solution strengthening of the ferritic structure and fine pearlite structures. These steels are normally used at the room, and sometimes at low, temperatures.

However, in the case of the high strength low- alloy medium carbon steel, it is usually based on either (Ni- Cr- V) or (Cr- Mo- V/W). The high strength of these steels comes about as a result of tempered alloyed martensite as well as the secondary hardening carbide precipitates. The structure of these steels in the heat treated condition permits high strength at higher temperatures usually up to 300-500 ℃, Robert (1967).

Alloy design for ultra high strength steels depends on high alloy systems based on strengthening by solid solution, dislocation and precipitation mechanisms created by the formation of tempered martensite. This is the case of the maraging steels which employ the presence of high Ni, Co and Mo in an extremely low carbon steels, Decker (1963). These make them applicable for both low and high temperature conditions in the range -70℃ and 550℃, Pickering (1978). Maraging steels are not only strong, they are extremely tough.

Alloy design for high-carbon steels is based on the addition of carbide-forming elements such as Cr, Mo, W and V. These steels depend on the formation of alloy martensite and alloy carbide structures. These, in addition to the presence of high carbon content, are responsible for the high hardness properties both at low and high temperatures. Required toughness can be induced by an appropriate tempering process.

The application of alloy high carbon steels is then primarily for obtaining high hardness for the necessary wear resistance and high

softening temperature, Grossman and Bain (1964). These are the general requirements for most tools steels which are derived from high carbon steels.

2.1.3 Classification Of High Carbon Steels

High carbon steels, alloyed or unalloyed, are generally used for the manufacture of tools. Hence they can be adequately contained in the various classes of tool steels with carbon content in excess of 0.65%. In this respects, the AISI classification of tool steels is adopted. This work concerns the use of carbon steels for tool making and excludes the applications requiring high strength and high elastic properties as are applicable in springs, wires, steels for ropes among others. The AISI list of tool steels completely contains all classes of high carbon steels used in the industry. The relevant portion of this list is produced in Table 2.1.

It is worthy of note that only few of commonly used tool steels in the AISI list do not contain carbon greater than 0.65%. These are:

(a) Class 220 – low alloy (<3% Cr) tool steels

(b) Class 240 – low alloy (<3% Ni) tool steels.

(c) Many classes of special– purpose tool steels in group 300.

(d) Group 500– hot work tool steels with carbon contents less than 0.65%

Table 2.1 AISI List Of Tool Steels With %C Greater Than 0.65%,Robert et-al (1961)

Group	Classes	Grade
100	All types	Carbon Steels
200	230	Cr- based low-alloy tool steel with < 3% Cr Ni - based low alloy tool steel with < 3% Ni
300	340	W - based low alloy finishing tool steel
	350	High carbon low alloy steels
	360	Semi high speed steels
400	410	Oil-hardening cold work die steel
	420	Air-hardening cold work die steel
	430	High-carbon Cr-based cold work die steels
	440	High carbon Cr-W based cold work die steels.
600	All types	High speed steels.

2.1.4 General Properties Of Steels For Tool Making

The major characteristics of steels for tool making are wear resistance, resistance to softening on heating and toughness. Machinability, safety and distortion during hardening are some times specified in certain tools, but they do not always come into play in every application. Consideration may be given to other factors such as depth of hardening, which is related to internal stress characteristics, and grain size specification at full hardness.

The wear resistance of a steel is known to be influenced by its hardness, but hardness alone is not itself a measure of wear resistance. Comparatively, carbon tool steels have low wear resistance while the tungsten-based high speed tool steels have extrememly high wear resistance, Wilson (1975). It is known that wear resistance of high speed steels is further increased by the addition of Cobalt, Payson (1962). In all grades of steels, wear resistance increases with % carbon, Dell (1978). However, there seems to be no direct relation between resistance to wear and resistance to softening at elevated temperatures. It is, however, known that a steel which is wear resistant often posesses good resistance to softening.

Toughness of a steel varies widely with carbon content, heat treatment, surface condition, size and temperature of operation, among many other factors. Toughness increases with the size of the tool but decreases with depth of hardening, Burns and Pickering (1964). In general, high carbon steels, in their normal heat treated condition have brittle- to-

ductile transition temperature (BDTT) well above the room temperature for notched samples. Toughness measurements by impact methods for these steels using notched samples may not be found useful, Steven (1959). It is found to be applicable for those steels with lower carbon content or high carbon steels in the unhardened conditions. Infact this method is commonly used for medium carbon steels rather than high carbon steels. For this reason, torsional impact methods have been found to be more appropriate in rating the toughness of hardened high carbon steels, Greene and stout (1940)

2.1.5 Selection Of Tool Steels

Cost and availability are the primary consideration in the process of selecting a steel in the AISI lists. However final choice must take into account other important requirements listed in Table 2.2 for common tool steel materials.

Judging from the cost consideration alone,tool steels in groups 100 and 200 are the most attractive. Their ratings, in respect of toughness, surface hardness are shown to be as good as any other. These are carbon and low alloy steels and are also found not to be susceptible to decarburization on hardening. The biggest draw backs are essentially in the areas of wear resistance, hot hardness, depth of hardening and dimensional stability on hardening. These are the ones which have the widest range of applicable working hardness. From the study of the hardening characteristics of the carbon tool steels in group 100 and group 200 it is quite clear that they are suitable for the production of

light tools commonly used in the normal workshop. Compositions of some of these steels are shown in Table 2.3 while the complete compositions for the low alloy grades are in Table 2.4.

It is quite common for the group 100 tool steels to contain small additions of V and Cr not exceeding 0.5%, since both have beneficial effects on carbon tool steels. At this level, vanadium suppresses the grain coarsening tendency of these steels while chromium promotes deep hardening characteristics, Bain and Paxton (1966).

Table 2.2 Properties Of Some Selected Common Tool Steels Robert et-al (1961).

Group Class	Cost Rating	Working Hardness (Rc)	Surface Hardness (Rc)	Hot Hardness Rating	Depth of Hardening Rating	Wear Resistance Rating	Toughness Rating
100(W)	1	55-65	65-67	1	S	2-4	3-7
211(L)	1	45-64	58-67	2	M	2-3	4-6
232(L)	1	45-62	58-63	2	M	3	6
340(F)	3	64-67	65-68	3	S	8	2-6
410(O)	1	55-62	61-64	3	D	4	3
420(A)	3	55-62	63-65	5	D	6	4
430(D)	3	55-64	61-64	6	D	8	1-2
440(A/D)	3	58-66	64-66	6	D	9	1
600(T/M)	1-5	63-68	63-66	8-9	D	7-9	1-3

High number rating means high response to the property concerned.

S - Shallow
M - Medium
D - Deep

2.1.6 Low - Alloy Tool Steels

The low-alloy high carbon steels in group 200 are Cr-or Ni-based (Table 2.1). Small quantity of chromium is added for deep hardening characteristics, and some quantities of molybdenum and vanadium for improved strength, toughness and wear resistance. These are used for the manufacture of tools such as shears, drills and blades of intricate shapes.

Table-2.3 Plain High Carbon Tool Steels Robert et-al (1961)

Types	%C	%Mn	%Si	%Cr	%V
110	0.60-1.40	0.25	0.25	-	-
120	0.60-1.40	0.25	0.25	-	0.25
121	1.00	0.25	0.25	-	0.50
131	1.00	0.25	0.25	0.50	-
140	1.00	0.25	0.35	0.35	0.25

Table 2.4-Low Alloy High Carbon Tool Steels Robert et-al (1961)

Types	%C	%Mn	%Si	%Cr	%Ni	%W	%Mo	% V
211(L_1)	0.65-1.10	<0.60	<0.5	075-1.50	-	-	-	-
232(L_6)	0.75	0.55	0.25	1.00	1.75	-	0.50	-
340 (F_2)	1.30	0.25	0.25	<1.00	-	3.5	>0.50	0.15
350(F_2)	1.25	0.30	0.30	0.35	-	1.45	-	0.20
410(O_1)	0.90	1.20	0.25	0.50	-	0.50	-	0.20
420(A_2)	1.00	0.60	0.25	5.00	-	-	1.00	-

The Ni-based grades, as in class 230, are the tough tool steel materials applicable for wood- working saws, knives, blanking dies and shears. Chromium is normally added for deep hardening and improved hardness of the case, while Ni strengthens and toughens the core.

Other low alloy tool steels such as classes 340 and 350 are W-base and have properties which include excellent wear resistance, high surface hardness both at the cold and hot condition. However, they are shallow hardening for which Cr in added to improve it. Applications of such steels are for cases of severe wear resistance such as hacksaw blades and wire drawing dies. The class 340 tool steels are harder but are prone to decarburization on prolonged heating at the normally high hardening temperatures.The class 350 tool steels are tough and strong because they do not contain enough W for secondary hardening during heat treatment.

The use of high Mn in low alloy tool steels, as in class 410 and class 420 tool steels is similar to that of Ni in class 220. However, this is accompanied with appreciable additions of Cr and Mo, as in the class 420 series, for deep hardening and additions of V and W are required for wear resistance in tools with keen cutting edge. These steels are prone to high retained austenite and hence require multiple tempering to remove it and to prevent the risk of brittleness.

2.2 Ductile Cast Iron

Ductile cast irons have graphite nodules embedded in a steel matrix. Depending on the heat- treatment it is subjected to, the matrix may be ferritic, ferrite-pearlite, pearlitic, bainitic or martensitic. Ductile cast irons have many attractive qualities which make them suitable for machine building and tool making. The greatest of which is its extremely high toughness, at the hardened conditions and its wear resistance in the presence of graphite nodules, Walton (1971).

Ductile cast iron cannot be used in applications requiring mechanical working such as forging, Schuyten (1975). Hence it is limited to direct use after casting, machining and heat-treatment. It is however included in this study because of the similarity in the structures and carbon content of the austenite of both ductile cast iron and high carbon steels at the hardening temperatures.

2.3.0 Heat Treatment Of Carbon Tool Steels

2.3.1 General

The plain carbon tool steels are among the least expensive of the tool steels and their properties include relatively low wear resistant, resulting in shorter tool life than those of alloy steels. Depth of hardening is shallow but forms tools with hard case and tough core. They are unsurpassed in performance in applications such as shear blades, blanking dies, reamers threading dies, taps, twist drills, wire drawing dies and wood working tools.

Carbon content in these tool steels ranges from 0.65%C. to 1.25%C. However according to Dell (1978), good combination of properties is obtained when the carbon content is between 0.8 and 1.10%. In the lower %C range there is reduction in the wear resistance though toughness is increased. For higher carbon levels, there are problems associated with hardenability, dimensional stability, tendency for cracking among others. In this range of carbon level, carbon tool steels are prone to the formation of cementite network at the prior austenite grain boundaries, especially for slow cooling from above the upper critical temperature through the critical range. This is very detrimental to the mechanical properties of these steels. Though the wear resistance of the hardened carbon tool steels may be enhanced by the cementite network, the toughness and machinability are drastically reduced. It is therefore necessary to remove any previously formed carbide network by a normalizing heat treatment, a prerequisite before the intended heat treatment.

2.3.2 Annealing Of High Carbon Steels

Annealing of high carbon hyper- eutectoid steels is performed essentially for two main purposes, Grossman and Bain (1964). One is to soften the steel for easy machining and the other is to produce a structure that will respond uniformly to hardening. Because of prolonged heating during the annealing operation, care should be paid in minimizing scaling and decarburization. In this case, they are packed in an inert substance such as dry sand. The use of packing materials of charcoal or cast iron chips is some how controversial since some decarburization has been reported particularly for annealing, Wilson (1975). The pack is then heated and soaked for 15 minutes/25mm size at an annealing temperature between 750-800 ℃. Cooling rate from this temperature should be slow, less than 25℃/hr down to 550℃, at which point fast cooling can be applied.

The structure commonly formed consists of spheroidised cementite in the matrix of ferrite. The micrographs below illustrate the various annealing structures for the 1.0%c plain carbon steel where:

(a) Fast cooling has resulted in pearlite structures.

(b) In sufficient slow cooling has resulted in annealing with very fine spheroidal and fine pearlite structure.

(c) Full annealing has produced fine spheroidal structures and

(d) Coarse Spheroidite is formed from full annealing.

In general, partial annealing is favoured for the hardening and machinability characteristics.

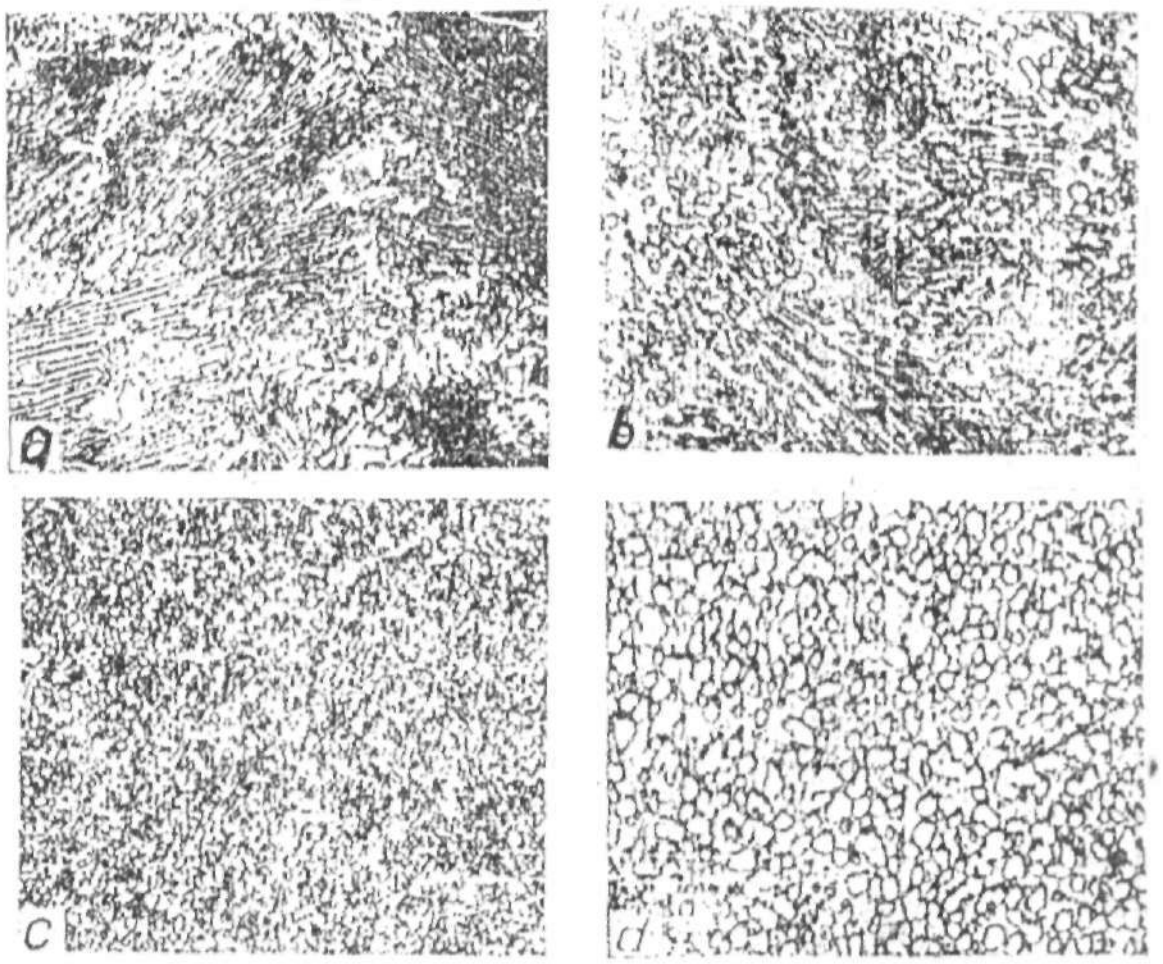


Plate 2.1: Microstructures of 1.0%C Plain Carbon Steel, x 500
Robert et-al (1961).

(a) First Cooling - Pearlite
(b) Partial annealing - Mixed structures of pearlite and spheroidal carbides
(c) Full annealing - Fine spheroidite
(d) Full annealing - Coarse spheridite.

2.3.3 Hardening Of High Carbon Steels

High carbon steels are commonly used in the hardened condition since wear resistance is one of the most important characteristics. In this respect, surface hardness of Rockwell (C65 to 70) is possible. This hardness reduces rapidly to the centre, as shown in Figure 2.1 for class 100 carbon tool steels. In this case Rockwell C55 hardness is normally regarded as minimum for hardened carbon tool steels, but Rockwell C50 is minimum sufficient for hardened medium- carbon steels. These correspond to about 50% martensite and 50% pearlite in the structure, Robert et-al (1961)

The actual hardness obtained after hardening will depend on the structure of these steels prior to quenching. This is with respect to the nature of the residual carbide present, composition and homogenisation of the prior austenite, Siebert et al (1977). All these are directly affected by the hardening temperature for a given high carbon steel.

The structure in the hardened zone of plain high carbon steels consists of martensite, residual carbide, not dissolved during soaking at the hardening temperature, and retained austenite, Wilson (1975). The relative quantities of these constituents determine the properties of the hardened steel and will be influenced by the following, Robert et al (1961):-

(a) Composition of prior austenite,

(b) Hardening temperature,

(c) Quenching procedure and cooling rate,

(d) Austenite grain size.

In short, all these factors are critical to the hardening characteristics of high carbon steels which produce a martensite- austenite-carbide mixture. Both martensite and carbide promote high hardening and good wear resistance, but retained austenite reduces it, Gardner (1956). The hardness of martensite increases with %C but, the hardness of hardened carbon steels levels out at 0.6%c due to the presence of retained austenite, Cohen (1949); Dell (1978), and begins to decrease when an appreciable amount of retained austenite in excess of about 20% level is present.

2.3.4 Effect Of Hardening Temperature

The hardening temperature of the hyper-eutectoid steels will influence the structure and composition of austenite before quenching. For ordinary high carbon steels, the hardening temperature is limited to within the critical range but it is very extensive and exist to much higher temperatures for alloy steels which form stable carbides, Roberts and Mehl (1943); Smith (1959).

The maximum hardening temperature normally corresponds to that at which full homogenization of austenite is achieved. Heating to higher temperature in excess of this, results in the formation of coarse grain structure in the prior austenite. Plain carbon steels are sensitive to rapid grain growth which occurs even at temperatures within the critical range, Payson (1962). Partial growth inhibition occurs with the addition of Cr for which the grain coarsening temperature is raised. Once grain growth starts in these Cr-steels, it grows very rapidly from

that temperature and above. Addition of strong carbide formers such as Nb ,V ,W and Mo, is very effective in suppressing grain coarsening in steels,as is illustrated in the Figure (2.2).

In the case of alloyed steels,the austenite becomes more enriched in carbon and alloying elements at high temperatures, and this causes an alteration in the transfomation characteristic of the steel, Smith(1959). The most important effect of this concerns the lowering of the M_s-M_f temperatures and widening of the M_s-M_f range. These and other effects are reflected in the reduction of there hardenability.

In the final analyses, low austenitizing temperatures cause poor homogenization, high residual carbide and questionable hardened properties. On the other hand, excessively high austenitizing temperature will cause high alloy and carbon concentrations, high retained austenite, dimensional instability of hardened parts, cracks and failures.

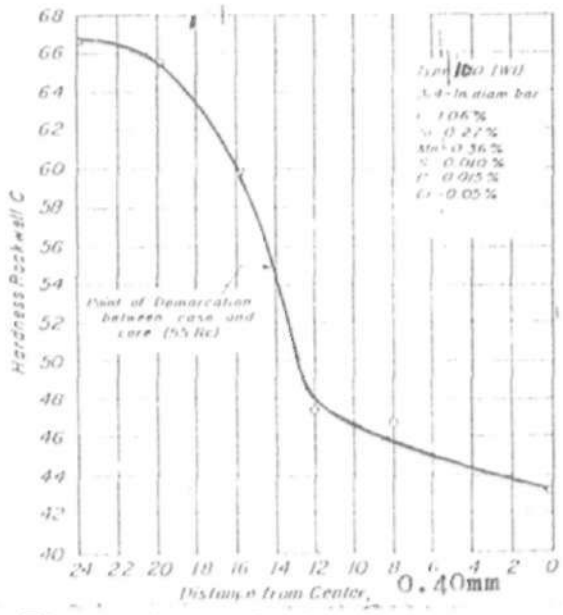


Figure 2.1: Hardness-penetration of 20mm round bar of type 100(W) quenched from 815°C.
Robert et-al (1961).

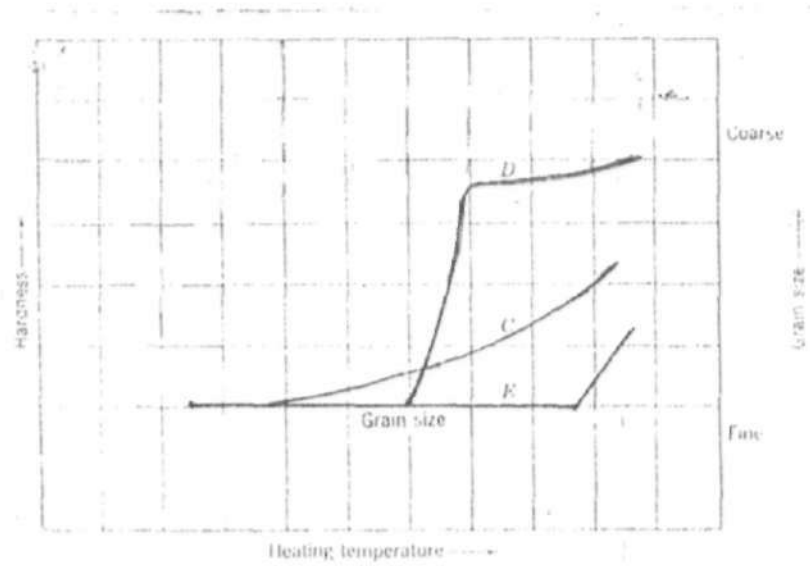


Figure 2.2: Schematic diagram showing Correlation of austenitizing temperature with maximum attainable hardness and with austenite grain size.
Payson (1962).

2.3.5 Hardenability Of Carbon Steels

The hardenability study of high carbon steels is not as simple as one might expect. For a given grain size, and when no residual carbide is present,it has been shown that, as the carbon content of a hypo-eutectoid plain carbon steel increases, hardenability continuously increases, Digges (1938). However, for the hypereutectoid carbon steels this is not the case, infact the eutectoid steelhas been shown to be the most deep hardening, Foley (1938). The presence of small quantities of S, Mn and Si in commercial high carbon steels is known to have been responsible for the wide scatter in the depths of hardening observed for these steels. Barrow and sholer (1943).

The austenitizing temperature has been shown by Wilson (1975) to be very important in the hardenability of high carbon steels.
Firstly, the austenite composition is dependent on it and the presence of high carbon in austenite, in excess of 0.8%C, means that there will be lowering of the M_f temperature range to below the room temperature, which causes the formation of an appreciable amount of retained austenite and proness to quenching cracks, Arpi (1931). Secondly, the tendency for grain coarsening for the high carbon steels is high at the hardening temperatures, Siebert et al (1977). Coarse grain structure retards austenite transformation at temperatures greater than 450℃ but does not matter at temperatures less than 450℃. This increases hardenability, causes formation of large martensite plates and promotes quenching cracks or tharmal shock, Digges (1938).

Thirdly, the structure of hyper-eutectoid steels, prior to hardening influences their hardenability.

For a given austenite composition, the presence of residual carbides will further reduce the hardenability of a given steel. In this respect, the deep hardening characteristics of a hyper-eutectoid steel will depend on both the composition of austenite prior to hardening and the size and distribution of the residual carbides, Gross man and Bain (1964).

Normalizing as a pre-hardening treatment, results in fine pearlite structures which dissolves fast on heating and causes fine dispersion of residual carbide. The presence of fine carbide particles in austenite has a greater tendency of causing a rapid nucleation of pearlite during austenite - transformation than the presence of coarse carbides. For high carbon steels close to the eutectoid steel, as in 1%C steel, normalizing treatment prior to hardening at the normal hardening temperature of about 800℃ causes practically no residual carbide to form. This hardening temperature is close to the critical temperature for grain coarsening. Hence for good deep hardening of these steels, normalizing prehardening treatment has been recommended, Payson (1962).

Spheroidizing annealing before hardening ensures small prior austenite grain size and coarse residual carbide at the hardening temperature. As a prehardening treatment, sphroidising annealing will reduce the hardenability of these steels, Grossman and Bain (1964).

For high carbon steels with carbon content in excess of 1%C, residual carbides are always present at the hardening temperature, and these cause reduced hardenability. The presence of residual carbide at the hardening temperature and high retained austenite(γ) on hardening have a complex inter-relationship which has been found to cause low hardenability in these steels, Payson and Klein (1943). It has also been found, Payson (1962), that it is advantageous to carry out a spheroidising prior hardening of this grade of carbon steels. Table 2.5 clearly explains this.

Table 2.5 Effect of prior hardening condition on the hardenability of some high carbon steels. Payson (1962)

Steel	Depth of case(mm) at hardening temperature							
	800℃		825℃		850℃		900℃	
Composition. (%C)	OQ	A	OQ	A	OQ	A	OQ	A
0.85	7.94	4.37	9.53	5.93	9.53	9.53	9.53	9.53
1.14	2.77	3.18	3.18	3.97	3.18	3.97	4.76	4.76

OQ = Oil quenching

A = Annealing

2.3.6 Retained Austenite In Quenched Carbon Tool Steels

Carbon Tool steels have low M_s and M_f temperatures, and these decrease with an increase in the carbon level. In theory, M_f temperatures fall below the room temperature for greater than 0.60% ℃ in carbon steels but the M_s temperature will not be sub-zero until %C>1.2, Amtsberg

(1945). In practice, the M_s-temperature for the 1.2% C steels is generally observed to be above the room temperature because of the presence of the residual carbide at the austenitizing temperature. However, the M_f temperature for this steel, is certainly below the room temperature,and full martensite transformation cannot occur except for a sub-zero treatment or quenching, Parker (1945).

The presence of retained austenite(γ) in quenched high carbon steels is not desirable not only because of its influence on hardenability, but also its influence on the:

(a) Tempered structure and properties,

(b) Dimensional stability,

(c) Ageing characteristics at the room temperature,and

(d) Cracking or brittleness in service of components not properly tempered.

The amount of retained austenite(γ) is very much related to the M_s and temperature of quenching medium. In this respect, Kointinen and Marburger (1959) developed a relationship for calculating the volume of retained austenite in quenched steels, and is

$$V_\gamma = \exp[(-1.10)\ x\ 10^{-2}(M_s - T_q)] \qquad (1)$$

$$M_s > T_q > -80\,^{\circ}C$$

V_γ = Volume fraction of retained austenite γ.

T_q = Temperature of the quenching medium °C)

M_s = Martensite start temperature (°C).

It is established by Cohen (1949), that higher alloy contents, higher austenitizing temperatures, slower cooling rates through the M_s-M_f temperature range and longer austenitizing times tend to result in greater quantities of retained austenite in the as quenched structures.

Averbach and Cohen (1949) suggested that quenching cracks observed in tool steels is caused by the ageing reactions occurring at the room temperatures due to the presence of retained austenite. The stabilization of retained austenite for interrupted quenching within the M_s-M_f range, seems to confirm this thinking, Harris and Cohen (1949). Hence it is always advised that retained austenite in tool steels should not be allowed to age after quenching.

2.3.7 Tempering

Tempering of quenched hardenable steels is a must and is done to achieve various objectives. In high carbon steels, tempering is done primarily to relieve the internal stresses generated during quenching and to maintain structural and dimensional stability for a maximum attainable hardness. To achieve these, tempered martensite is formed and any retained austenite present must be transformed, Lement et al (1953).

Transformation of retained austenite is accompanied by a high volumetric expansion within a martensitic structure. When the quantity of retained austenite is small, the volumetric expansion resulting from the austenite transformation may easily be contained at most of the tempering temperatures. However, when the amount of retained

austenite is appreciable and is transformed at the tempering temperature, the volumetric expansion caused by the transformation of retained austenite may be so large that a brittle untempered martensite may not be able to contain it. This is the cause of temper cracks which embrittle these steels, Averbach and Cohen (1949). This problem is commonly observed when tempering high carbon steels, and is not common with carbon steels with less than 0.6%C. Deliberate efforts must therefore be made to reduce the amount of retained austenite present before tempering high carbon steels. This is not only because of the danger of temper brittleness but also because it affects the as-quench hardness.

There is always a temperature range associated with this type of temper brittleness. Tempering at low temperatures will not cause austenite transformation. However, tempering at higher temperatures will cause the prior martensite to temper and soften even before austenite transformation occurs. In the latter case, the martensite can deform to accommodate the associated volumetric expansion during the subsequent austenite transformation.

For purposes of wear resistance applications, carbon tool steels are tempered between 150 and 250℃. In this temperature range, transformation of retained austenite is limited, and so also the formation of tempered martensite structures. Quenched 1% carbon steels, tempered in this temperature range have been shown to produce a loss in plasticity in torsional stress-strain curves, Greene and Stout (1940).

The same results were obtained by Palmer and Luerssen (1948) in special torsion impact test for which toughness peaks were obtained at specific tempering temperatures, Figures 2.3 and 2.4. These are temperatures at which optimum tempering conditions have been found to occur, Greene and Stout (1940). Beyond these temperature levels, toughness of these steels, on tempering begins to drop. This phenomenon is known as 500F embrittlement, Capus and Mayer (1960); Klinger et al (1954) For a given steel, the temperature at which this toughness peak occurs is found to vary with the hardening temperature, Figure 2.4. Palmer and Luerssen (1948) reported that it varied between 150 and 215℃ for 1.00%C steels hardened between 750 and 800℃.

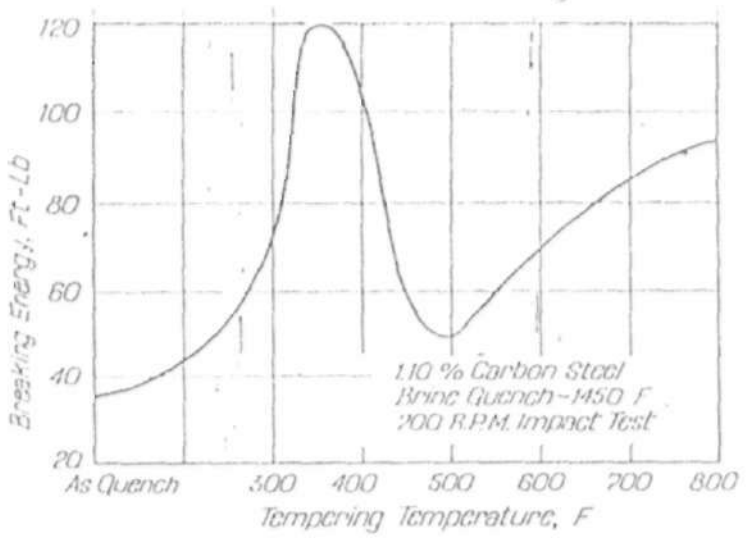


Fig. 2.3: *Effect of tempering temperature on torsion impact values of quenched 1.10% C tool steels* (Palmer and Luerssen)

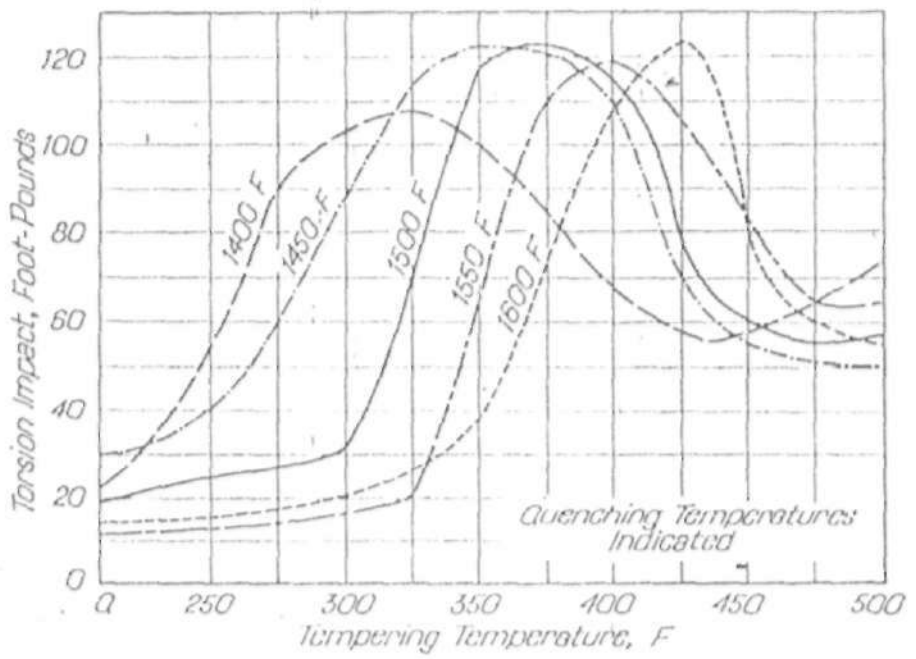


Fig. 2.4: *Effect of quenching temperature on the relation between torsion impact strength and tempering temperature for 1% C tool steel* (Palmer and Luerssen)

Gill and Bowman (1923) found that impact test on unnotched quenched and tempered high carbon steels do not show these toughness peaks, and the toughness values continuously increased with the tempering temperatures. Also the toughness values for these steels, tested under the notched impact conditions, were found to be too low and were some what erratic.

For high applications requiring considerable toughness, a high tempering temperature in excess of 300℃ was common for carbon tool steels. In this case, tempered martensite with stable cementite is formed, and all retained austenite will be transformed to lower or upper bainite. Beyond 350℃ quenched plain carbon tool steels soften rapidly and greater than 50 Rc is hardly attainable, Holloman and Jaffe (1945).

It is important to note that all high carbon steels when quenched from the tempering temperature suffer from reduced toughness, Gill and Bowman (1923). This may be due to the formation of additional internal stresses. Thus slow cooling from the tempering temperatures is recommended. In a similar note, long time tempering at low temperatures, has been found to increase the strength, toughness, hardness and elastic properties of these steels.

CHAPTER 3

3.0 EXPERIMENTAL PROCEDURE

3.1 Introduction

A study aimed at evaluating the hardening characteristic of certain high carbon materials which were hardened and tempered, was carried out.Materials and methodology of investigation is shown below.

3.2 Materials

Five grades of high carbon materials were studied. These consisted of:-

(a) Three plain high carbon steels with %C between 0.8 and 1.2.

(b) One low-alloy high-carbon steel material and

(c) One ductile cast iron.

The chemical composition of these materials are contained in table 3.2. It can be seen that specimens A and B are near 1%C carbon steels, and specimen C is at the 1.2% carbon limit for hard and tough steels. The low alloy high carbon steel was type 353(F) W-base finishing tool steel material.

The plain high carbon steels were obtained from the Department of Mechanical Engineering, and were supplied as rods in the hot-rolled condition and of diameters shown in Table 3.1. The low-alloy high carbon steels was obtained from the Electricity meter manufacturing company (EMCON) Nigeria limited based in Zaria. It was supplied as a

rod in the hot-rolled and cold-finished condition. The ductile cast iron samples were supplied in form of charpy impact test specimens.

Table-3.1 Materials Used

Materials			As-received dimensions (mm)	Dimensions of Test Specimens (mm)	
	AISI Class	Code		Diameter	Length
Plain Carbon Steel	200	A	23 dia	18	15
Carbon Steel	200	B	23 dia	18	15
Steel	200	C	20 dia	18	15
Low-alloy Steel	353	D	12 dia	12	10
Ductile Cast Iron	-	E	50 x 10 x 10		

Table 3.2 Chemical Composition Of Materials.

Materials		Chemical Composition Of Materials Used									
		%C	%Mn	%Si	%Cr	%V	%W	%P	%S	%Mg	%Ni
Plain . Carbon Steels	A	0.95	0.25	0.25	-	-	-	0.03	0.03	-	-
	B	1.05	0.30	0.25	0.25	-	-	0.03	0.03	-	-
	C	1.20	0.30	0.25	-	-	-	0.03	0.03	-	-
Low-alloy Steel	D	1.20	0.25	0.20	0.20	0.20	1.20	0.03	0.03	-	-
Ductile Cast Iron	E	3.63	0.36	2.26	0.07	-	-	0.03	0.01	0.01	0.05

Composition above is from the Manufacturers.

3.3. Methodology

The study of the hardening characteristics of these high carbon materials was based principally on hardness measurements of hardened specimens. To obtained some reasonable control of the experimental process and parameters, the following measures were taken.

1. The plain carbon steels specimens were machined to the same size

specimens were machined square.

2. One flat surface of each specimen was ground smooth using 240 - to 400 - grit papers. This made surface preparation for hardness measurement easy, especially after hardening.
3. All heat-treatment operations were carried out in a carbolite muffle furnace which can maintain furnace temperature within ±5℃. All heat treatment specimens were packed in carburizing media which consisted of cast iron chips mixed with pulverized charcoal,all were contained in a fire clay ceramic covered container.
4. Heating of the specimens for austenization was done, firstly by heating to 650℃ at the normal heating rate of the furnace. This was followed by slow heating from 650 and soaking for thirty minutes at 750℃. Final heating to the austenitizing temperature was rapid and this was followed by soaking for an appropriate time. Discharging the austenitized specimens into the quenching medium was done with minimum loss of temperature.
5. All hardening process was preceeded by a normalizing heat-treatment.

The following operations were carried out in the course of this study:-

(a) Normalizing heat treatment of all samples as a pre hardening treatment.

(b) Hardening at various temperatures and using water and oil as quenching media.

(c) Hardening at 800℃ and tempering at varying temperatures between 150 and 500℃.

(d) Hardening at 800℃, followed by refrigeration treatment and tempering at various temperatures.

(e) Impact test of the steel specimens which had been hardened at 800℃ and tempered between 200 and 500℃.

(f) Macroscopic examination of impact fractured surfaces of the specimens.

(g) Jominy end-quench test of specimens prepared from the plain carbon steel materials.

(h) Metallographic preparation and examination of specimens.

3.4 Pre-Hardening Normalizing Heat-Treatment

Twelve specimens were cut from each material and were prepared as in section 3.3 and packed inside a carburizing medium contained in a fire clay cup. There were sixty specimens altogether. The cup (and its content) were charged into the heat treatment furnace, and was heated to and austenitized at 900℃ for 1 hour. The cup with its content was then removed from the furnace and the specimens discharged to cool in air. Two sets of specimens from each material, were removed for metallographic observation of structures. Results are shown in the micrographs.

3.5 Hardening Heat Treatment

The remaining ten sets of specimens, which had previously been normalized in 3.4, were packed into ten covered graphite crucibles in a carburising enviroment and charged into the furnace.
In this respect, each crucible contained a set of five specimens from the five materials. The specimens were heated to and austenitized at 750℃ for 30 minutes. After, this time, one set of specimens was quenched in water and another set was quenched in oil.

The furnace temperature was raised to 800℃, and the charged materials were soaked for a further 30 minutes. Another set of specimens was quenched in water and one other quenched in oil. This process was repeated for specimens austenitised at 850, 900 and 950℃.

For each specimen, the smoothened surface, as prepared in section 3.3, was again lightly ground on a grinding paper using the 400 and 600 grit papers and this was finally followed by light polishing on the polishing wheel. The hardness of each specimen was measured on the polished face using the Avery model hardness testing machine. The results are shown in Table 4.1.1., 4.1.3 and 4.1.4., as well as Figures 4.1 and 4.2.

3.6 Tempering Characteristics

Eight sets of five specimens each, which had also been previously normalized, were prepared as in 3.3, were packed in the ceramic container and were austenitized at 800℃ for 30 minutes. The specimens were all quenched in water at room temperature. After the necessary surface preparations, the as-quenched hardness values were measured

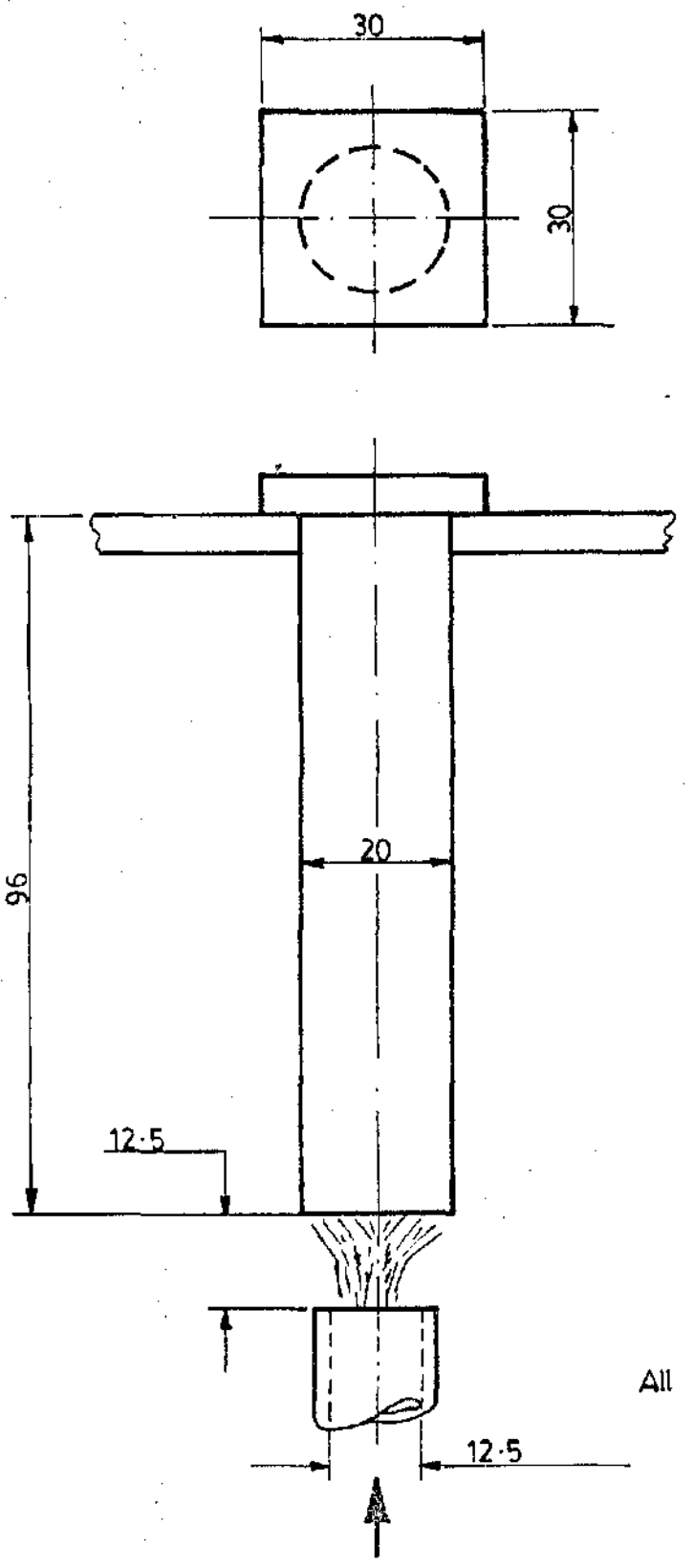
for one set of five specimens. Another set was packed in a carburizing medium in graphite crucible, tempered for 30 minutes at 150℃ and finally air cooled. The hardness values were measured again.

Seven other sets were packed in seperate crucibles, and from these a set was tempered at one of 200, 250, 300, 350, 400, 450, and 500℃. The results are shown in Table 4.1.5 and Figure 4.3.

3.7 Jominy End-Quench Test

Jominy end-quench test was carried out using only the plain high carbon steels. It was not possible to machine jominy end-quenched test specimens from the low-alloy steel and the cast iron materials. Because all the steel materials had diameters less than the standard 25mm diameter size and for the purpose of comparing hardenability, these steels were all machined to 20mm-diameter sizes. This is the diameter of carbon steel materials with the smallest diameter. The jominy end-quenched test specimens were packed for heat treatment, charged into the furnace heated to and autenitised for 1 hour at 800℃.

By the use of a dummy specimen, the water jet in the jominy end-quenched testing equipment was properly adjusted. One by one, the heated specimens were transferred into the appropriate support, and allowed to cool to the room temperature under the influence of water jet.A 0.4mm-deep flat was machined longitudinally on the surface of the specimens. At regular intervals hardness measurements were made along the length of the machined flat surface. The results are presented in Table 4.1.7 and Figure 4.5.



All Dimension in mm.

Jet of water at 27°C ± 2°

FIG. 3·1 Jominy End — Quench specimen .

3.8 *Impact Tests*

Seven izod impact-test specimens were made for each of the steel materials. These made twenty eight impact test specimens altogether. Each specimen, as shown in Figure 3.2, had two notches machined on it to facilitate two impact test measurements.

These impact test specimens were also given a pre-hardening normalizing heat treatment as for all the previous specimens. They were again packed in a carburing medium in a ceramic container, austenitised at 800℃ for 1 hour, and then quenched in water. A set of four specimens, one for each grade of steel, was tempered for 30 minutes at each of seven temperature intervals between 200 and 500℃.

Each specimen was tested twice by Izod impact blows and the average of the two izod impact test values was calculated. The results are shown in Table 4.1.8 and Figure 4.6.

3.9 Effect Of Cold Treatment

Another collection of eight sets of specimens,were prepared as in section 3.3, austenitized at 800℃ for 30 minutes quenched in water then ice water. They were immediately transferred into a deep freezer at -22℃ and kept for three days. For a set of specimens, the as quenched hardness values were measured as well as after tempering at 150℃ for 30 minutes. Other sets were tempered at temperatures up to 500 ℃. Their hardness values were also measured. The results are presented in Table 4.1.9. and Figure 4.7.

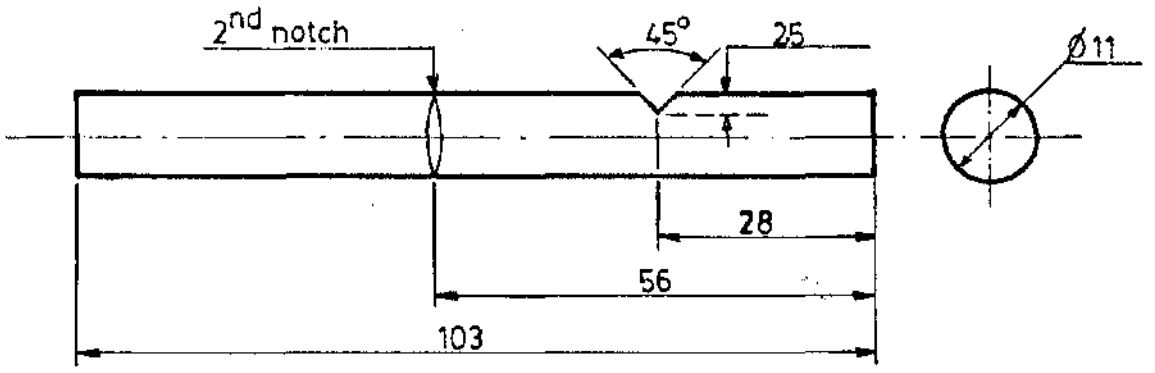
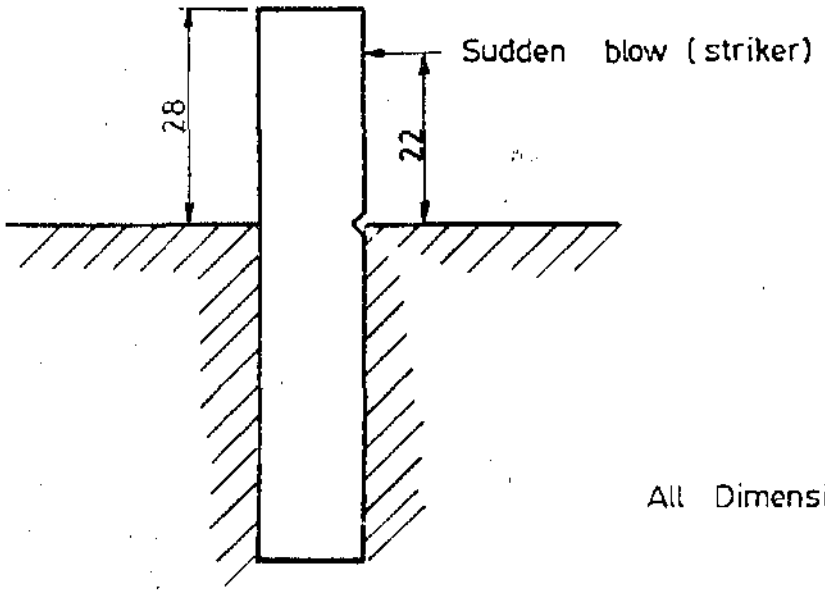


FIG. 3·2(a) Impact test specimen containing two V - notches.



All Dimension in mm

FIG. 3·2 (b) Izod test specimen and position of impact .

3.10 Metallographic Structures

Some samples from the test specimens were prepared for the observation of structures, and the following structures were of interest.

(a) As received structure,

(b) Normalized structure,

(c) As-hardened,

(d) Hardened and tempered.

Metallograhic preparations of samples had been made easy by the earlier pre-hardening grinding in 3.3 and previous polishing of some of the specimens for hardness measurements. However, for the purpose of metallographic observations of structures each specimens was carried again through successive fine grinding on strip grinders using 240, 320-400-and 600 grit papers. Care was taken to rotate the specimens by 90° when changing from one grade of paper to the next one. Also during grinding, water was generously added to ensure adequate cooling of the specimens. After the grinding was completed, the metallographic sample was rinsed with water before polishing.

Polishing was carried out on a rotating wheel using nylon cloth and 1μ aluminum powder. The surface of each sample was cleaned with cotton wool soaked in alcohol and etched with 2% Nitric acid in alcohol solution (Nital) for thirty seconds. The specimen was then rinsed in alcohol to stop the reaction of Nital, and was dried in air. The etched specimen was then observed under the microscope. Results are presented in micrographic plates 4.1 – 4.4.

CHAPTER 4

4.0 RESULTS AND DISCUSSION

4.1.0 *Results*

Table 4.1.1 Hardness of values (HV) Of Specimens Quenched In Water At Various Temperatures

Temperature(℃)	Specimens				
	A	B	C	D	E
750	454	515	390	511	150
800	925	899	779	939	211
850	940	957	833	910	383
900	925	964	850	844	498
950	859	855	797	888	471
As-received Hardness Value(HV)	398	483	372	497	144

Table 4.1.2 Residual Carbide At The Hardening Temperature

Quenching Temperature (℃)	Acementile Composition %C	% Residual carbide		
		A	B	C
750	0.87	1.20	2.70	5.00
800	1.02	-	1.20	2.70
850	1.17	-	-	1.20
900	1.32	-	-	-
950	1.45	-	-	-

Computed from Iron-Carbon phase diagram.

Table 4.1.3 Evidence Of Cracks In Water Quenched Specimens

Cracking Temperature(℃)	Specimens				
	A	B	C	D	E
800	NC	NC	NC	NC	NC
850	NC	C	NC	NC	NC
900	C	C	C	C	NC
950	C	C	C	C	NC

Table 4.1.4 Hardness Values (HV) Of Specimens Quenched In Oil At Various Temperatures

Oil quenched Temperature (℃)	Oil quenched Hardness (HV)				
	A	B	C	D	E
750	280	266	273	219	LV
800	343	328	336	315	184
850	400	395	355	816	261
900	421	356	334	823	423
950	568	432	408	856	423
As-received Hardness Value(HV)	398	483	372	497	144

LV = Low value.

Table 4.1.5 Hardness Values(HV) For Specimens Quenched At 800℃ In Water And Tempered For Thirty Minutes.

Tempering Temperature in ℃	Water Quenched at 800℃ Hardness values (HV)				
	A	B	C	D	E
As Quenched	781	744	720	660	378
150	785	762	600	619	382
200	756	758	653	601	371
250	718	720	582	521	367
300	643	695	520	501	360
350	576	611	489	474	354
400	518	531	420	459	354
450	446	429	387	468	345
500	319	290	260	464	343

Table 4.1.6 Comparison Of The As Quenched Hardness Values (HV) Of Test Materials For Figures 4.1 And 4.5 all Quenched From 800℃ In Water

Materials	A	B	C	D
As quenched Hardness (HV) in figure 4.1	925	899	779	931
As Quenched Hardness (HV) in figure 4.5	781	744	720	660
Difference in the as quenched Hardness(HV)	144	155	59	271

Table 4.1.7. Jominy- end Quenched Test Result

Distance from Quenched end 2mm units	Hardness of water quenched end at 800℃ (HV)		
	A	B	C
1	803	762	730
2	708	612	501
3	403	425	391
4	362	403	379
5	360	402	376
6	363	404	372
7	365	401	373
8	375	393	370
9	380	382	374
10	388	358	361

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Table 4.1.8. Results Of Impact Test Based On Izod Principle Conducted On Heat Treated Specimens Quenched At 800℃ And Tempered At Different Temperatures For Thirty Minutes.

Tempering Temperature in ℃	Specimens Water Quenched at 800℃, Fracture energy (J)			
	A	B	C	D
200	11.53	8.82	ND	23.05
250	10.17	17.29	9.16	23.73
300	10.50	24.92	20.68	28.48
350	14.24	39.66	31.87	32.54
400	11.53	35.26	26.11	20.34
450	7.45	18.41	25.43	32.54
500	8.14	17.17	26.45	51.53

J = Joules
ND = not determined.

Table 4.1.9. Hardness Values Of Specimens Quenched From 800℃ into ice Water, Refrigerated For Three Days And Tempered For Thirty Minutes.

Tempering Temperature in (℃)	Specimens Water Quenched at 800℃, Refrigerated Hardness values (HV)				
	A	B	C	D	E
As Quenched	792	814	815	896	402
150	804	823	817	897	401
250	829	817	804	804	392
300	715	746	736	610	384
350	690	661	680	715	378
400	652	639	666	661	377
450	557	527	511	598	379
500	493	487	462	571	364

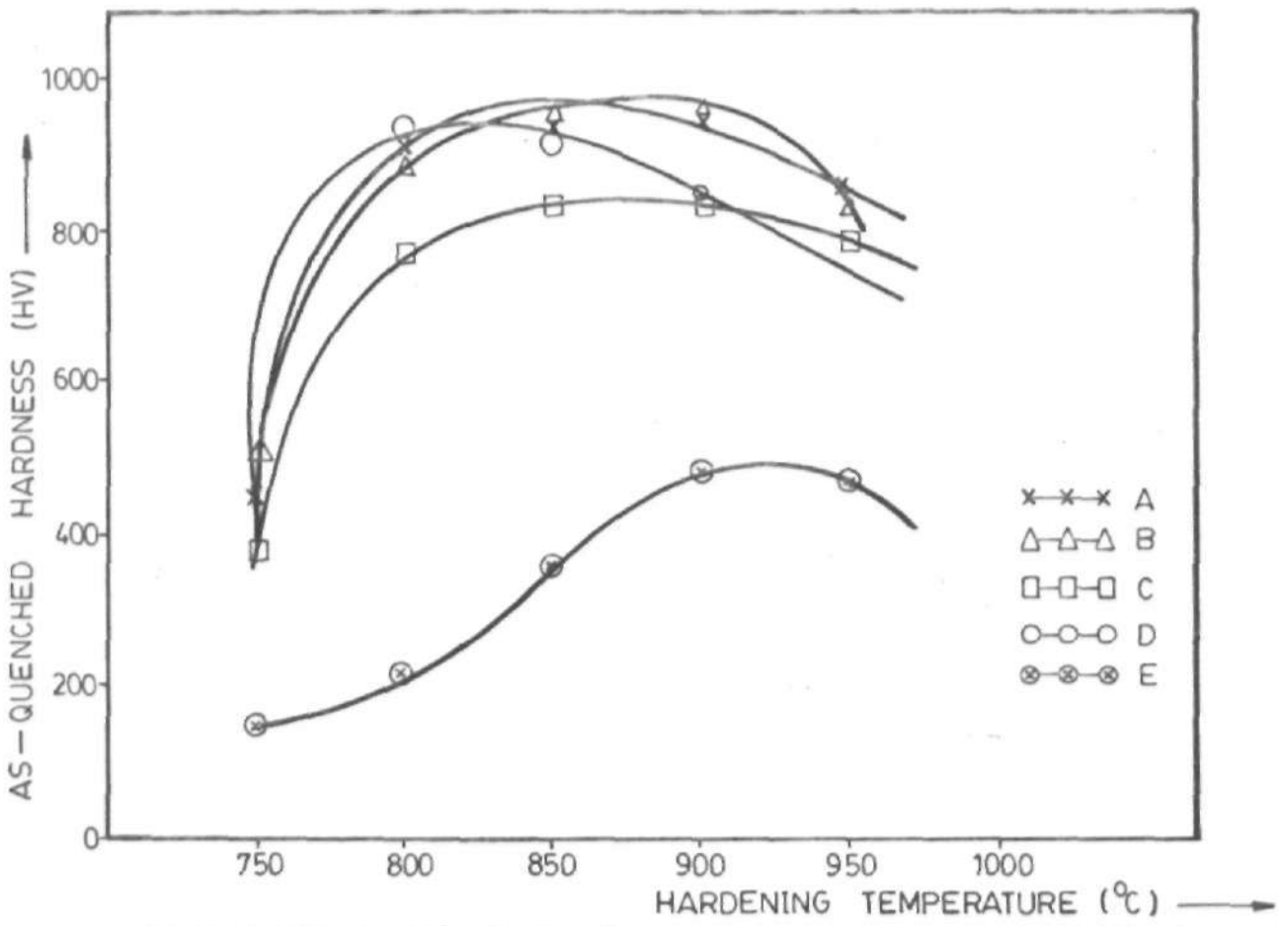


FIG. 4·1 : Effect of hardening temperature on the hardness (HV) of specimens, quenched in water .

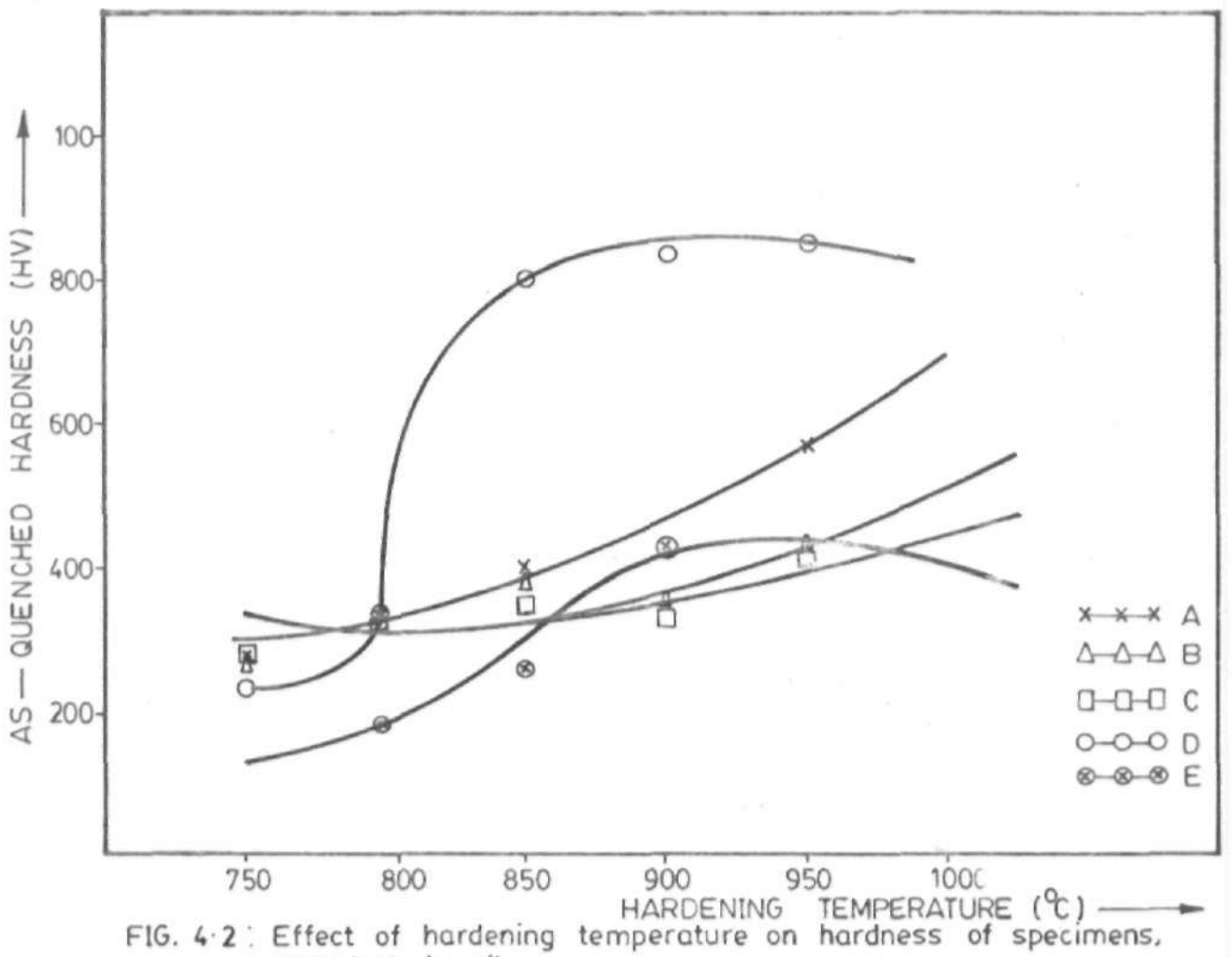


FIG. 4·2 : Effect of hardening temperature on hardness of specimens, quenched in oil .

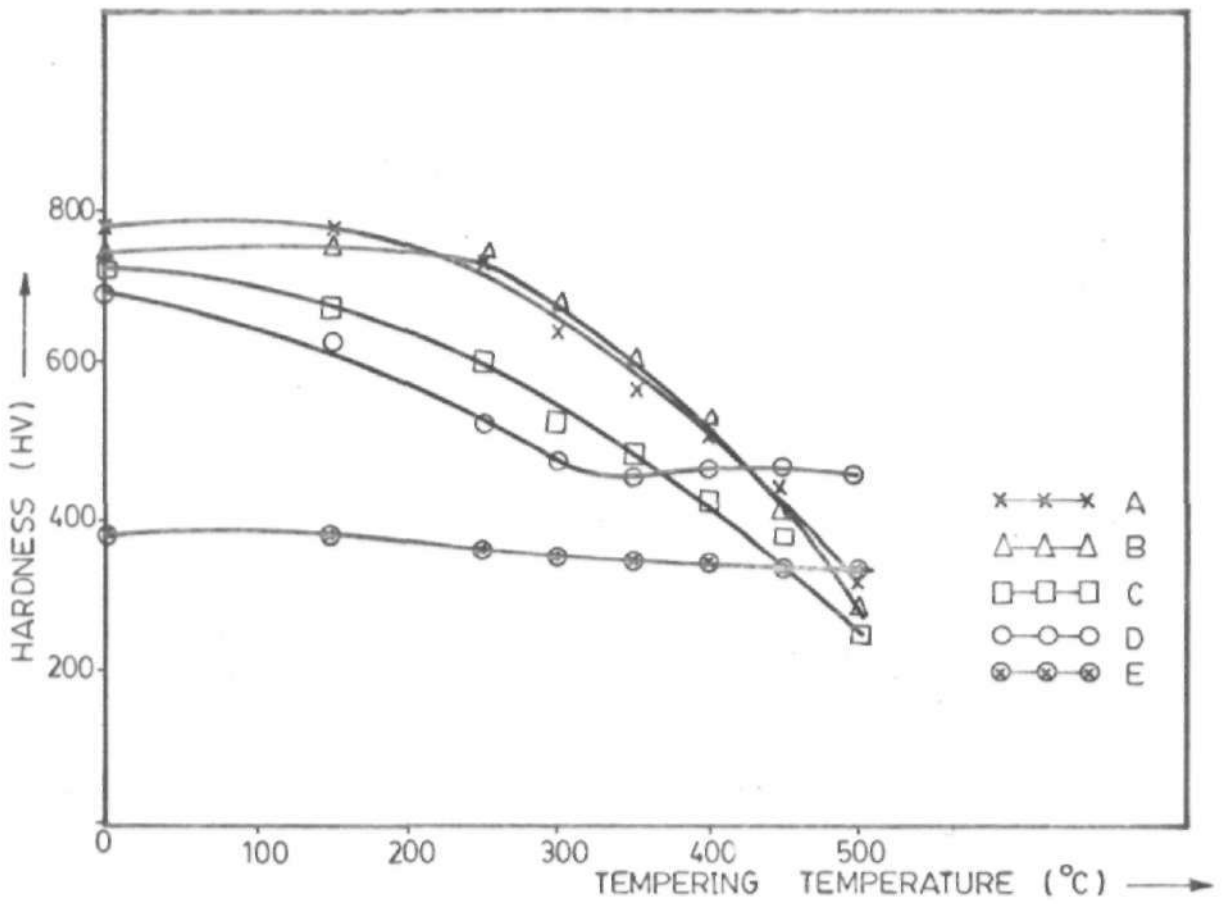


FIG. 4·3: Effect of tempering temperature on hardness (HV) of specimens quenched in water at 800°C, tempered for thirty minutes.

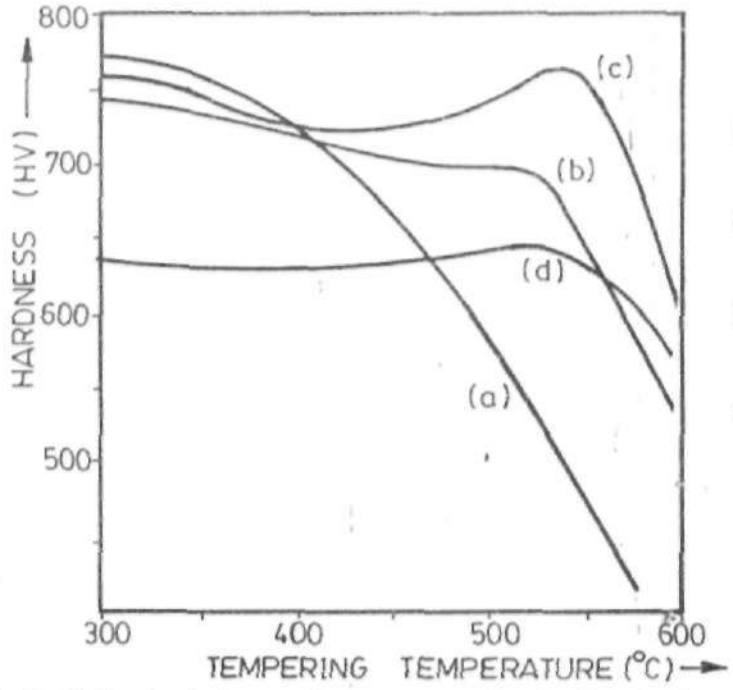


FIG. 4·4: Typical tempering curves for tool steels. Wilson (1975)

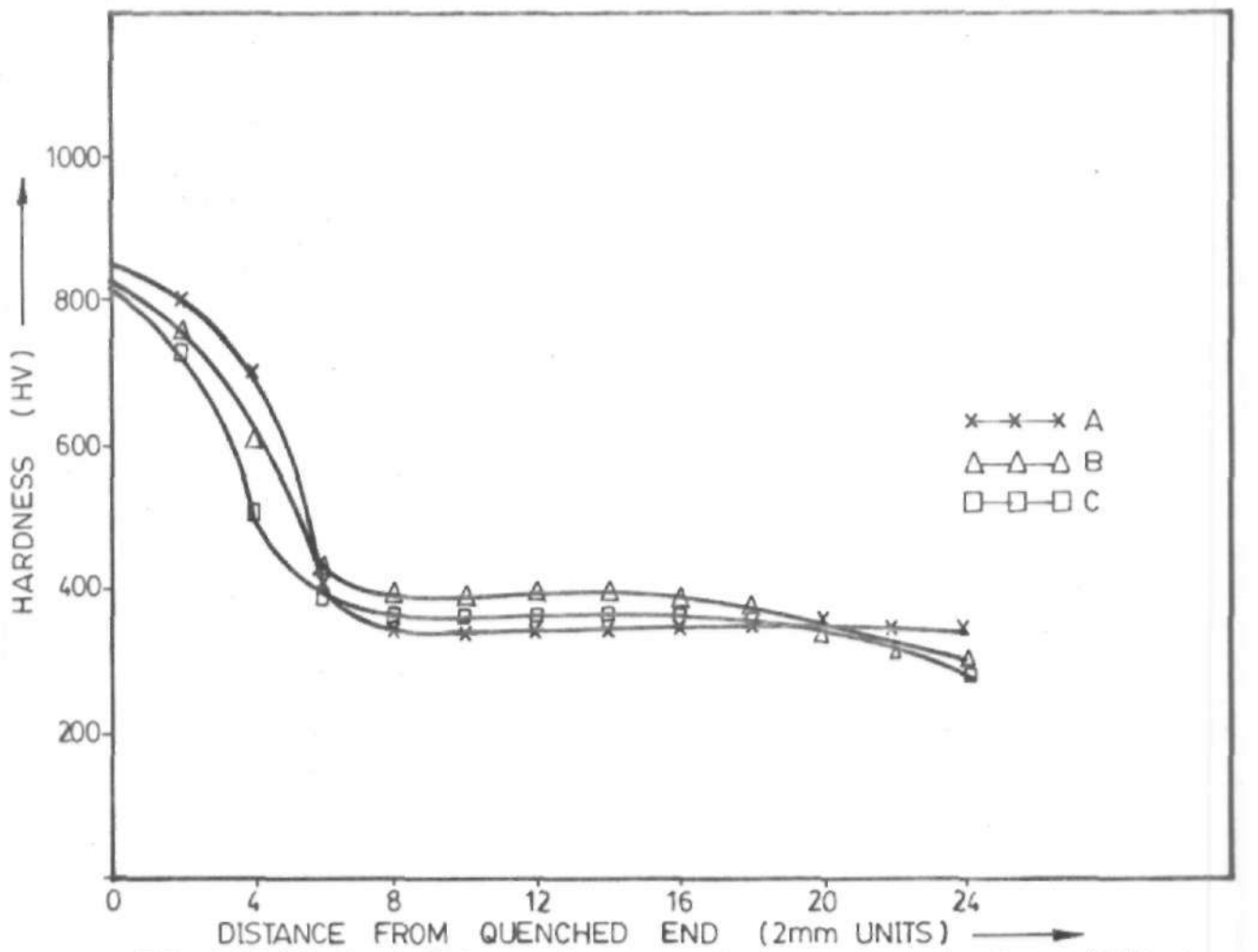


FIG. 4·5: Effect of distance from quenched — end on hardness (HV) of specimens quenched at 800 °C in water at the end .

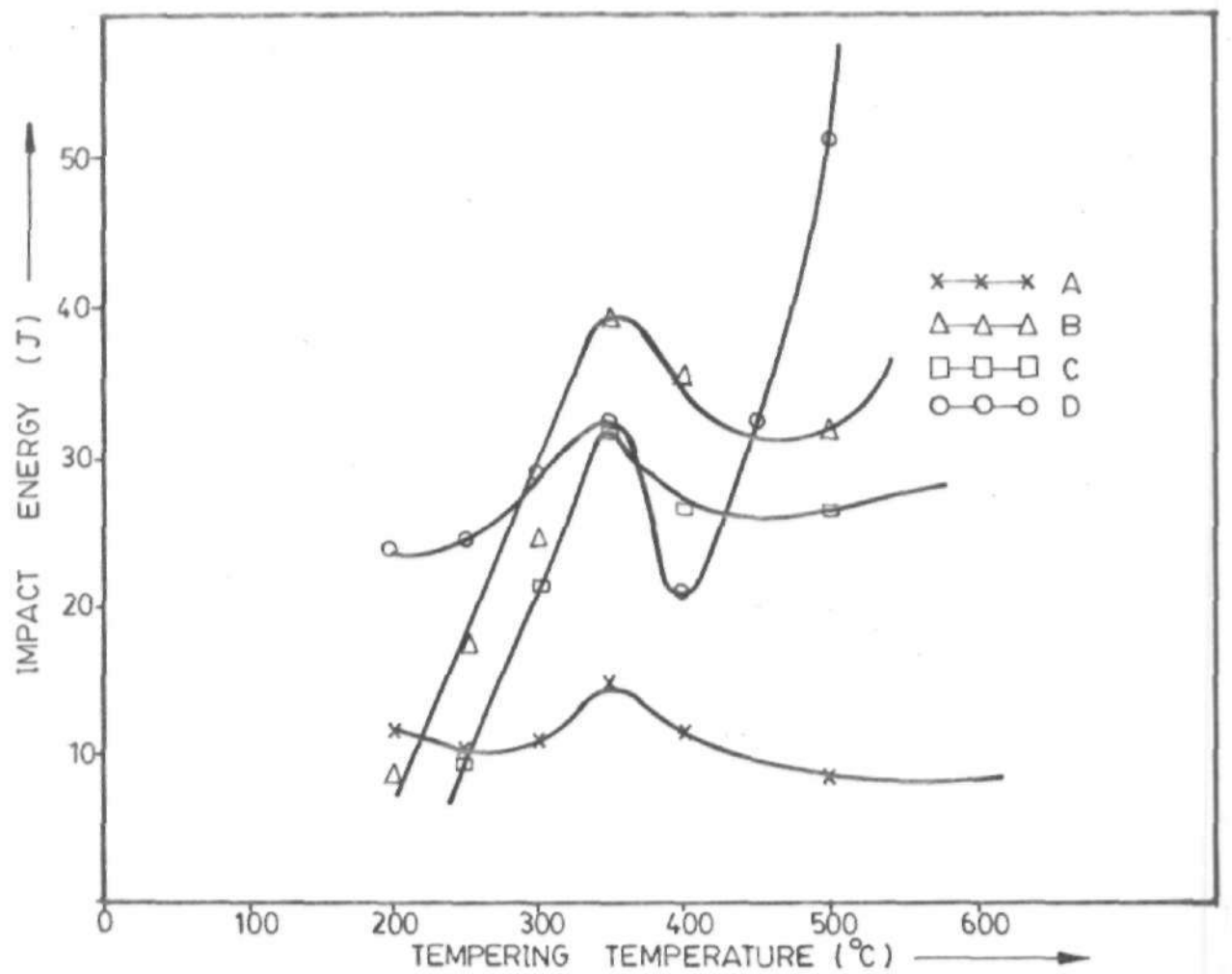


FIG. 4·6 : Effect of tempering temperature on impact energy of specimens quenched in water at 800 °C , tempered for thirty minutes .

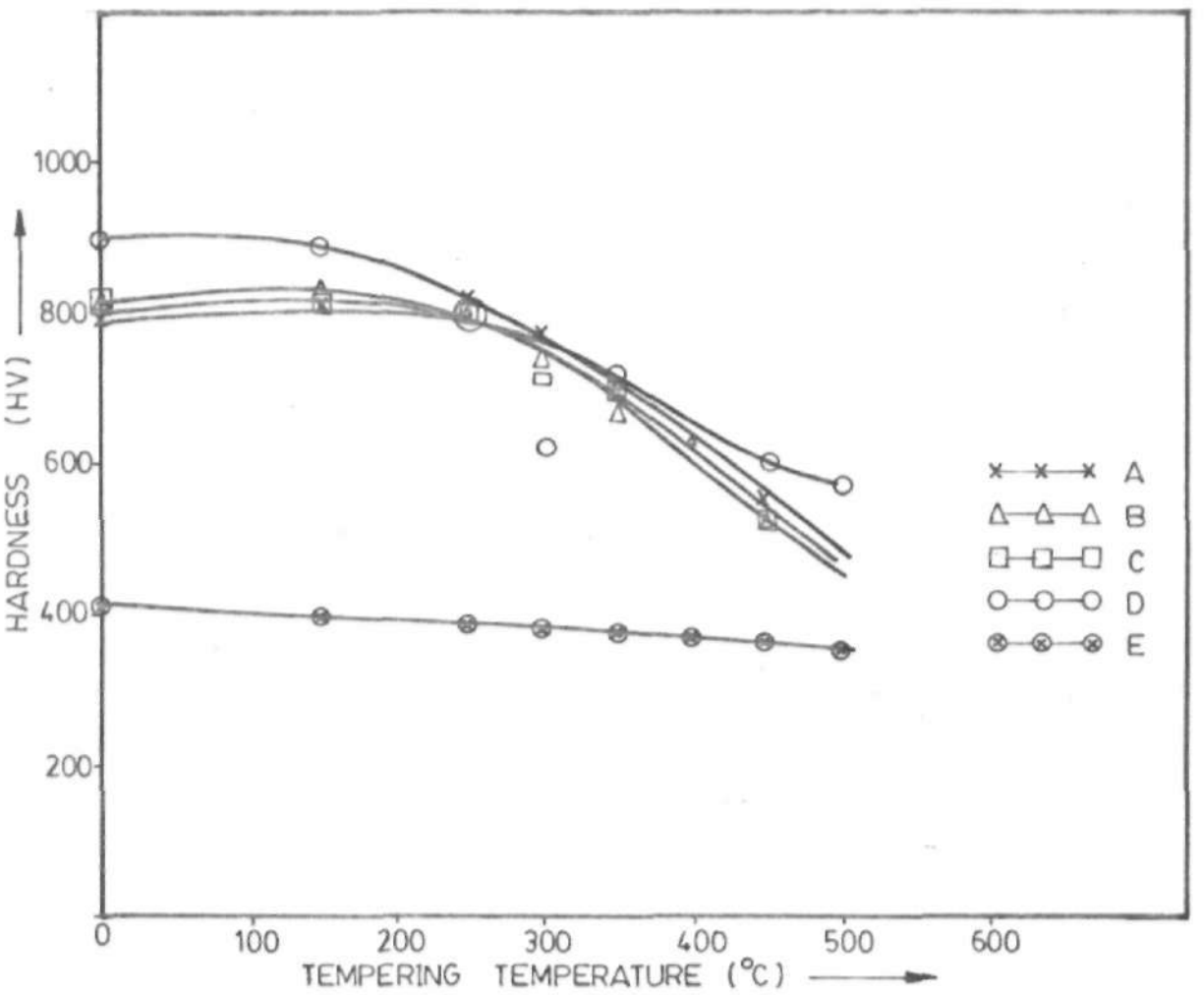


FIG. 4·7: Effect of tempering temperature on hardness of specimens quenched at 800 °C in water cold treated for three days, tempered for thirty minutes .

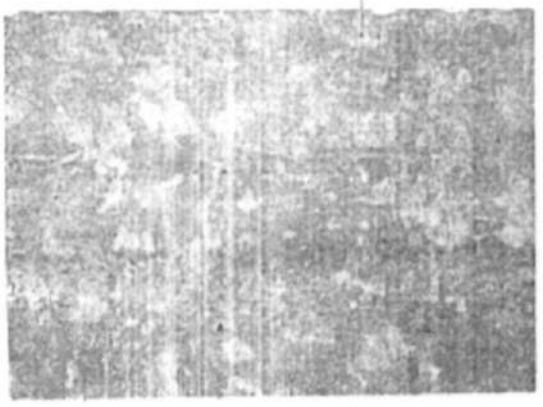


Plate 4.1(a)

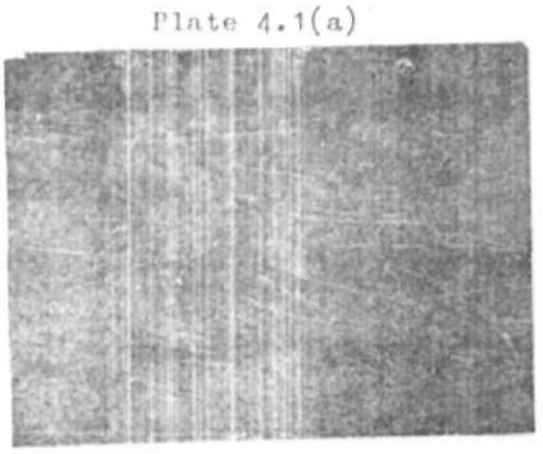


Plate 4.1(b)



Plate 4.1(c)

As-received Structures 500x

Plate 4.1(a) - 0.95%C and 1.05%C Plain Carbon Steel.
(Lamellar pearlite)

Plate 4.1(b) - 1.20%C Plain Carbon Steel.
(Spheroidal Carbides)

Plate 4.1(c) - As-cast Ductile cast iron.
(Graphite nodules in a ferrite matrix)

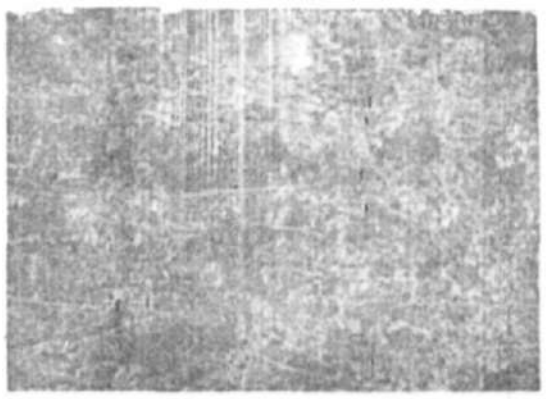


Plate 4.2(a)

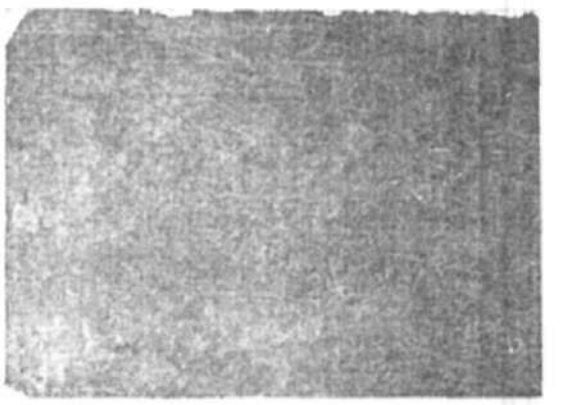


Plate 4.2(b)



Plate 4.2(c)

Normalised Structures 500 x

Plate 4.2(a) - 0.95%C, 1.05%C and 1.20%C plain carbon steel. (fully pearlite)

Plate 4.2(b) - 1.20%C low alloy steel. (Extremely feine precipitates of carbides)

Plate 4.2(c) - Ductile Cast Iron. (Pearlite Structure)

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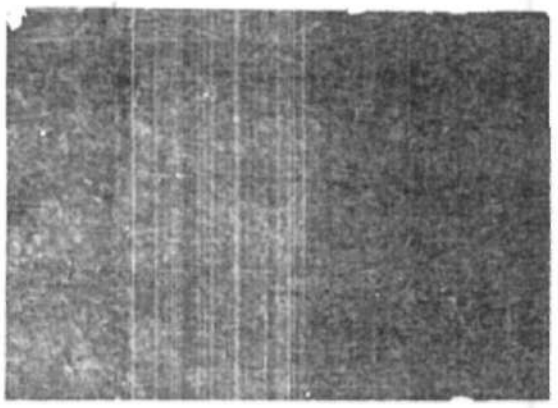


Plate 4.3(a)



Plate 4.3(b)

As Quenched Structures 500x

Plate 4.3(a) - All steel specimens
(Fine structures, presumably a mixture of martensite, fine bainite and pearlite.)

Plate 4.3(b) - Ductile Cast Iron.
(ferrite matrix)

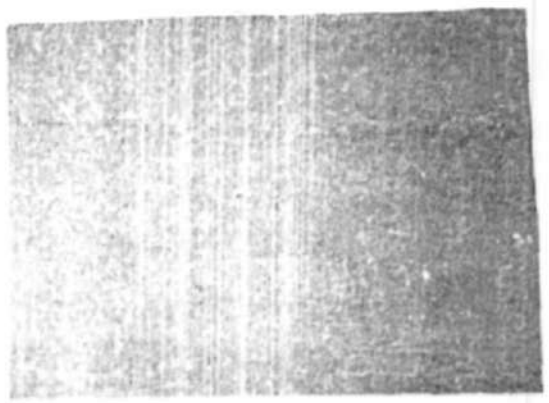


Plate 4.3(c)

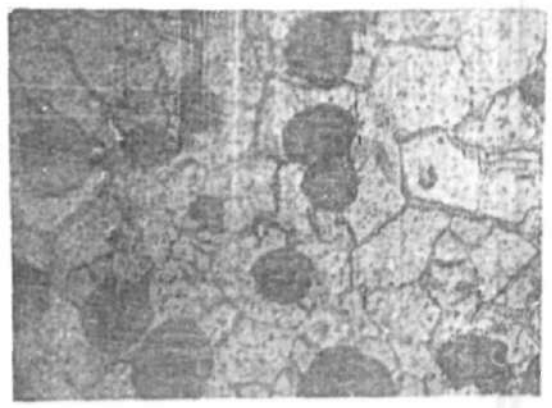


Plate 4.3(d)

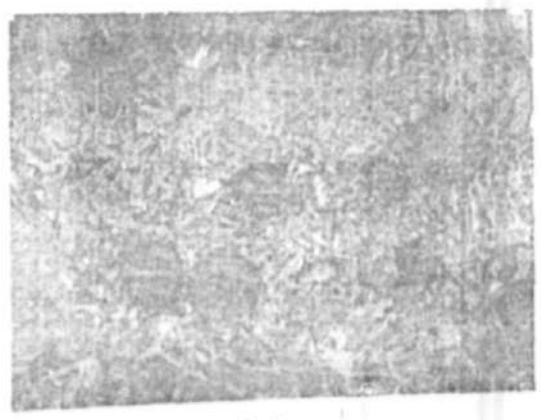


Plate 4.3(e)

Cold Treated Structures 500x

Plate 4.3(c) - All steel specimens.
(Extremely fine structures)

Plate 4.3(d) - Ductile Cast iron.
(fine structure)

Plate 4.3(e) - Ductile cast iron, tempered at 300°C.
(widmanstatten structure in a matrix of ferrite)



Plate 4.4(a)



Plate 4.4(b)

Impact Structures 300x

Plate 4.4(a) - 0.95%C Plain Carbon Steel.
(Martensite and bainite)

Plate 4.4(b) - 1.05%C; 1.20%C; 1.20%(C,W); Steels.
(Fine structures as in 4.3(a).

4.2.0 Discussion

4.2.1 As-Received Condition

The as-received structures for the high carbon steels consited of fine pearlite in all cases. However, while the 0.95%C and 1.05%C grades had lameller pearlite the 1.20%C had spheroidized carbide. These are shown in micrographs 4.1(a) and 4.1(b) respectively. The as-cast struture of the ductile cast iron was essentially ferritic with well distributed graphite nodules in a matrix of ferrite. There were scattered pearlite colonies as shown in micrograph 4.1(c).

4.2.2. Prehardening Treatment

Micrographs in plates 4.2a, b, c, give the normalizing prehardening treatment structures of the steels and ductile iron at x300 magnification.

The normalised structures were fully pearlitic, for the plain carbon steels, with no intergranular cementite precipitation, Micrograph 4.2(a). However, the normalised structure for 1.20%(C,W) low-alloy steel had extremely fine structure not resolved in the microscope with extremely fine precipitates of carbides which were visible in micrograph 4.2(b). Normalised ductile cast iron specimen was completly pearlitic, micrographs 4.2(c).

4.2.3 Effect Of Hardening Temperature

Table 4.1.1. shows the as-quenched hardness values for the high carbon steels and ductile iron material hardened by water quenching. Figure 4.1, however relates these hardness values with the

austenitizing temperature for the water quenched condition. In this figure, it is observed that the as-quenched hardness increased with the hardening temperature, but reached a maximum level at some temperature which depended principally on the composition of the steel. These are the optimum hardening temperatures if hardness is the only concern, and are between:

(a) 800 and 900℃ for the plain carbon steels,

(b) 800 and 850℃ for the W-based low-alloy high carbon steels,

(c) 900 and 950℃ for the ductile cast.

Hardening at higher temperatures resulted in some loss of hardness, even when the hardening temperature was above the Acem. point. However, when hardening was done below the optimum temperature, it resulted in formation of unhardened structures, which were caused by the presence of residual carbides, and incomplete homogenization of austenite. Table 4.1.2 shows the amount of residual carbides calculated for the various hardening temperatures.

The maximum as quenched vicker's hardness were found to be about 950, 970, 850 and 950 (HV) for the 0.95%C, 1.05%C, 1.20%C carbon steels, and the low-alloy steel respectively. The 1.20%C carbon steel was found to have the lowest hardness of the steels, and this was expected since the hadenability of carbon steels decreases as the carbon content increases in the hyper-eutectoid steels. Retained austenite, in the as-quenched condition, has been reported to be the chief cause, Cohen (1949); Wilson (1975).

The hardness of water-quenched low-alloy highcarbon steel was found to be quite high when compared with the 1.20%C carbon steel, which had the same carbon content. This clearly illustrates the alloying effects on the hardenability and possibly the wear resistance of the 1.20%C carbon steel.

The hardness values of the ductile cast iron were very low. The highest value, shown in Table 4.1.1., for water-quenching was 498 (HV) (50Rc) and which was attained at a high temperature of 900℃. This is barely the hardness for lower bainite or structures with 50% martensite. Ductile cast iron has very low hardenability, which may be attributed to the effects of very fine graphite particles as precipitates, seedlings and nodules. Silicon content may also contribute to the reduction of hardenability of ductile cast iron.

Practically all steel specimens quenched in water from 850℃ and above, suffered from quench cracks, Table 4.1.3. However, no cracks were found in all the ductile cast iron specimens. In this respect hardness values are not the only consideration in choosing the austenizing temperature. For hypereutectoid carbon steels, hardening temperature must be limited to below 850℃, Robert et al (1961).

Table 4.1.4 and figure 4.2 contain corresponding hardening data for oil quenching for the same materials. From these, all the plain high-carbon steels, and the ductile cast iron specimens failed to attain full hardening. Practically all hardness values were less than 500 HV (50Rc) in all cases. The hardness values for the ductile cast iron specimens

were particularly low being less than 450 HV (45Rc) and which could be the hardness of fine pearlite. For the oil quenching conditions, all the specimens did not suffer from quench cracks, and this was understandable and expected.

The low-alloy steel component was found to be hardenable by oil quenching from 850℃ and above. Hardness values of over 800 HV (65Rc) were observed. These are significantly higher than the hardness values obtained for the plain high-carbon steels, and clearly shows the effectiveness of small additions of strong carbide forming elements for increased hardenability in high carbon steels, Smith(1959). It was also found that the steel did not suffer from quench cracks even when oil quenched from 950℃.

4.2.4 Tempering Characteristics

Tempering caused softening of hardened structures in all the steels studied in this work as shown in Table 4.1.5 and Figure 4.3. In the case of the plain high carbon steels, water quenched from 800℃, there was an initial gradual reduction in hardness for tempering at temperatures up to 250℃. Beyond this temperature, softening became more rapid. Tempering at 500℃ seemed to cause a stable structure with hardness of 300 HV (30Rc) for all the carbon steels tested. The low-alloy high carbon steel showed the same initial gradual reduction in hardness with tempering temperature, but softening was arrested for tempering temperatures above 250℃.

In all these cases the initial reduction in hardness was probably as a result of stress relief in these steels. It should be remembered that it has been found that these steels are prone to high residual stresses on water quenching, Greene (1930). However the rapid softening, suffered by the plain carbon steel after tempering at higher temperatures could be due to the precipitation of stable incoherent cementite precipitates. Spheroidization of cementite must have stated as tempering temperature approached 500℃. These observations are characteristics of plain carbon steels.

Softening arrest observed in the case of the low-alloy steels is also characteristic of steels alloyed with strong carbide forming elements. This is due to formation of fine dispersion of coherent precipitates in tempered martensite or bainite. These also cause the characteristic secondary hardening in some steels tempered at higher temperatures, usually, above 500℃. Common tempering curves for different steels are shown in Figure 4.4.

Ductile cast iron, which was quenched in water, did not show any changes in hardness values when tempered at temperature less than 500℃ covered in this work. The hardness on tempering remained constant at about 360HV (36Rc) which corresponds to that of fine pearlite in carbon steels. Tempering of pearlite or bainite below 500℃ does not result in any structural changes but softening of this structure may only be observed when stress relief occurs. However, this was not the case in ductile cast irons.

It is noteworthy that the as-quenched hardness values for the steels in Figure 4.3 were substantially lower than the corresponding ones in Figure 4.1. Table 4.1.6 is extracted from Tables 4.1.1 and 4.1.5 to show that the difference was particularly significant for low-alloy high carbon steels, and least for the 1.20%C plain carbon steels.The data in Figure 4.1 were actually obtained some months after the heat treatment was performed, at that time the hardness measuring equipment was faulty and was not repaired until two months later. Ageing process, common in hardened steels with retained austenite, is suspected to be responsible for the marked increase in the observed as-quenched hardness values in figure 4.1. Avarbach and Cohen (1949) have vividly illustrated that austenite-martensite reaction in steels does not stop sharply immediately after quenching, but that it may continue at a diminishing rate for months. Infact ageing is what accounts for the dimensional instability observed in many tool steel.

4.2.5 Hardenability Measurements

The results of the jominy end-quenched test for the plain high carbon steels specimens are presented in Tables 4.1.7 and Figure 4.5. In Figure 4.5, 600 HV (55Rc) intercept was drawn on the jominy curve and the corresponding distances from the quenched ends of the jominy specimens were obtained. These distances were 4.8mm, 4.0mm and 3.5mm for the 0.95%C, 1.05%C and 1.20%C carbon steels respectively. By the use of an appropriate chart, these distances were converted to rough estimates of the critical diameters for these steels, and were 16mm, 14mm and 12mm respectively according to increasing carbon level. These sizes represented the hardenability values for the plain carbon

steel. That for the low-alloy steel was calculated using appropriate hadenability factors for the alloying elements contained, and it came to be 65mm. The implication of these is that these sizes of steels will harden through when quenched in water from 800℃.

4.2.6 Impact Tests Of Specimens

Notched izod impact test results for the specimens, which were quenched in water from 800℃ and tempered at various temperatures, are shown in Table 4.1.8 and Figure 4.6. The Figure shows that toughness peaks were obtained for these steels tested by notched impact tests. This is different from expectation since such test were reported to produce erratic results, Gill and Bowman (1923). The emergence of these peaks may be attributed to the structures which consisted of tempered martensite and possibly bainite formed from transformed martensite.It is striking to note that all the peaks were found at 350℃ tempering temperature. This seems to suggest that for all plain high carbon and low-alloy high carbon steels, the optimum toughness does not depend on the carbon content. The maximum toughness for these steels were 14J, 40J and 33J for the 0.95%C, 1.05%C and 1.20%C plain carbon and low alloy steels respectively. In general steels with higher carbon contents will be expected to give greater toughness values. However, a closer study will be needed to determine the inter-relation between carbon content and toughness on one hand as well as austenitizing temperature and optimum tempering temperature on the other hand.

Visual examination of fractured surfaces revealed a general fibrous appearance, and no case and core structure. However, macro and micro-examinations showed a mixed mode of fracture.

4.2.7 Cold-Treatment Of Specimens

The specimens which had been quenched in water from the hardening temperature of 800℃ were cold-treated in a deep freezer at -22℃, and tempered at various temperatures. The results are presented in Table 4.1.9. and Figure 4.7. In general the results are similar to those in Figure 4.3 However the following differences are significant:

(1) The hardness values were all higher except those for the ductile cast iron which remain unhardened even after cold treatment. The change, for the low-alloy steel specimens, was very remarkable and deserve comments. Infact this steel was found to be the most superior of them all, because of the high tempered and untempered hardness values. The hardness after cold treatment increased from a mere 650HV (58Rc), for no cold treatment to 900 HV (67Rc) with cold treatment. Also the high-temperature hardness values were 450HV (45Rc) and 580 HV (55Rc) respectively. This observation further confirms the presence of retained austenite in these steels after quenching in water, and is transformed either by cold treatment or ageing processes, Parker 1977. In addition, ductile cast iron contained no retained austenite to be transformed by cold treatment.

(2) All the plain high carbon steels developed virtually the same hardness values both in the tempered and untempered condition. This is evident in closeness of the tempering curves

for these steels. This seems to suggest a similarity in their hardening characteristics when room temperature quenching is followed by cold treatment. The conversion of retained austenite, after cold treatment, provided the same structure which consisted essentially of martensite.

4.2.8 Metallographic Examination

In the as-quenched condition without cold treatment, all the steel had very fine structures not resolvable in the light microscope, micrograph 4.3(a). These are presumed to be a mixture of martensite, fine pearlite and bainite structures. Retained austenite was not present due to ageing process, which could have converted it to lower bainite. The banite was probably extensive, overshadowing any martensite that would have been formed. The overall appearance was very much like tempered martensite. The corresponding as-quenched ductile cast iron was entirely ferritic with no pearlite colonies, micrograph 4.3(b). This shows that 800℃ is not the austenitizing temperature for the ductile cast iron, which is basically Fe-C-Si ternary alloy. This temperature is the temperature for feritizing treatment of ductile cast iron. This explains why the as-quenched hardness was low 400HV (40Rc) and also explains why tempering had no effect on ductile cast-iron quenched from 800℃.

When ice water quenching was used, followed by cold treatment, extremely fine structure as shown in micrographs 4.3(c) was obtained, for all the steels. Tempering at 300℃ did not produce any perceptible change under the light microscope. The same was observed for the quenched and cold treated ductile cast iron samples as seen in

micrograph 4.3(d). However, tempering at 300℃ produced a widmanstatten structure scattered in a matrix of ferrite, micrograph 4.3(e). This is not expected, and could not be explained.

Micrographs of sections cut from the izod impact specimens revealed no case-core structure, and microstructures for these samples were essentially as shown in micrographs 4.4(a and b) for which fine structures are as in 4.3(a).

CHAPTER 5

5.0 CONCLUSIONS AND RECOMMENDATIONS

5.1 Conclusions

The hardening characteristics of the five high carbon materials showed that:

1. Hyper-eutectoid plain carbon steels were hardenable by water quenching , but not by oil quenching. However, the low-alloy high carbon steel was hardenable by both oil and water quenching. Ductile cast iron was found to be barely hardenable by water quenching only, after heating to and soaking at high temperature.
2. Optimum hardening temperatures for these high carbon steels was found to be between 800 and 850℃. The actual temperature will depend on the size of the component. Hardening at temperatures above 850℃ certainly caused quench cracks in these steels. Hardening temperatures less than 800℃ did not cause full hardness.
3. Optimum hardening temperature for the ductile iron specimen was between 900 and 950℃. Outside this, range, full hardness will not be attained. It is remarkable that ductile cast irons are not susceptible to quench cracks when quenched from 950℃.
4. For a given hardening temperature, the as-quenched hardness values decreased as the carbon content increased. This was probably due to the presence of retain austenite and residual carbides.

the addition of 1.20%W into 1.20%C steel caused a further increase in the amount of retained austenite on quenching.

9. Optimum tempering temperature for maximum toughness was found to be 350℃ for all steels quenched from 800℃ and immediately tempered. This was irrespective of the composition of the high carbon steels covered in this work. Generally, because of the presence of retained austenite, the maximum toughness values for these steels increased with carbon content when no sub-zero treatment was performed. This is not expected in hardened high carbon steels for which a subzero - treatment had been done prior to tempering.

10. Critical diameters were 16mm, 14mm and 12mm respectively for 0.95%C,1.05%C and 1.20%C carbon steels. These clearly show the low hardenability values of these steels. However, it is 65mm for the 1.20%C, 1.20%W low-alloy steel by a rough estimate.

5. Hardening at 800℃ produced average hardness values of 823, 772, 633, 680 and 395 HV for the 0.95%C, 1.05%C, 1.20%C carbon steels, 1.20%C, 1.20%W low-alloy steel and cast iron respectively.These correspond to about 65, 63, 57, 59 and 41Rc respectively. This shows that ductile cast iron was not hardenable by quenching from 800℃.

6. Tempering immediately after quenching, at temperatures below 250℃ caused a slight reduction in the as-quenched hardness of the steels. This could be due to the relief of internal stresses present after quenching, and which were responsible for the characteristic quench cracks in these grade of steels.

7. At temperatures above 250℃ tempering caused rapid softening of the plain high carbon steels. However, softening was arrested in the case of the W-based low-alloy steel for which no softening occurred even up to 500℃ tempering temperature. This softening arrest was probably due to secondary hardening characteristics of alloying steels with strong carbide forming elements.

8. Conditioning treatment by cold treatment at -22℃ caused retained austenite in high carbon steels to transform into martensite. This then resulted in improved hardness of these steel both in the tempered and untempered condition. This was particularly true for both the 1.20%C plain carbon steel, and 1.20%C, 1.20%W low steel for which there was an appreciable amount of retained austenite after quenching. The increase was particularly significant for the latter which showed that

5.2 Recommedations

The hardening characteristics of high carbon steels and ductile cast iron, quenched in water and oil, at different temperatures, have not been adequately concluded.

1. Even though the optimun hardening temperature for plain high carbon steels was found to be at 800℃, further work is hereby, recommended with regards to the size effect.
2. Low-alloy high carbon steel seemed to be the most appropriate where toughness and hardness requirements are high, the oil-hardening characteristics should be investigated.
3. The response of ductile cast iron to hardening and tempering at higher hardening temperatures should be carried out.

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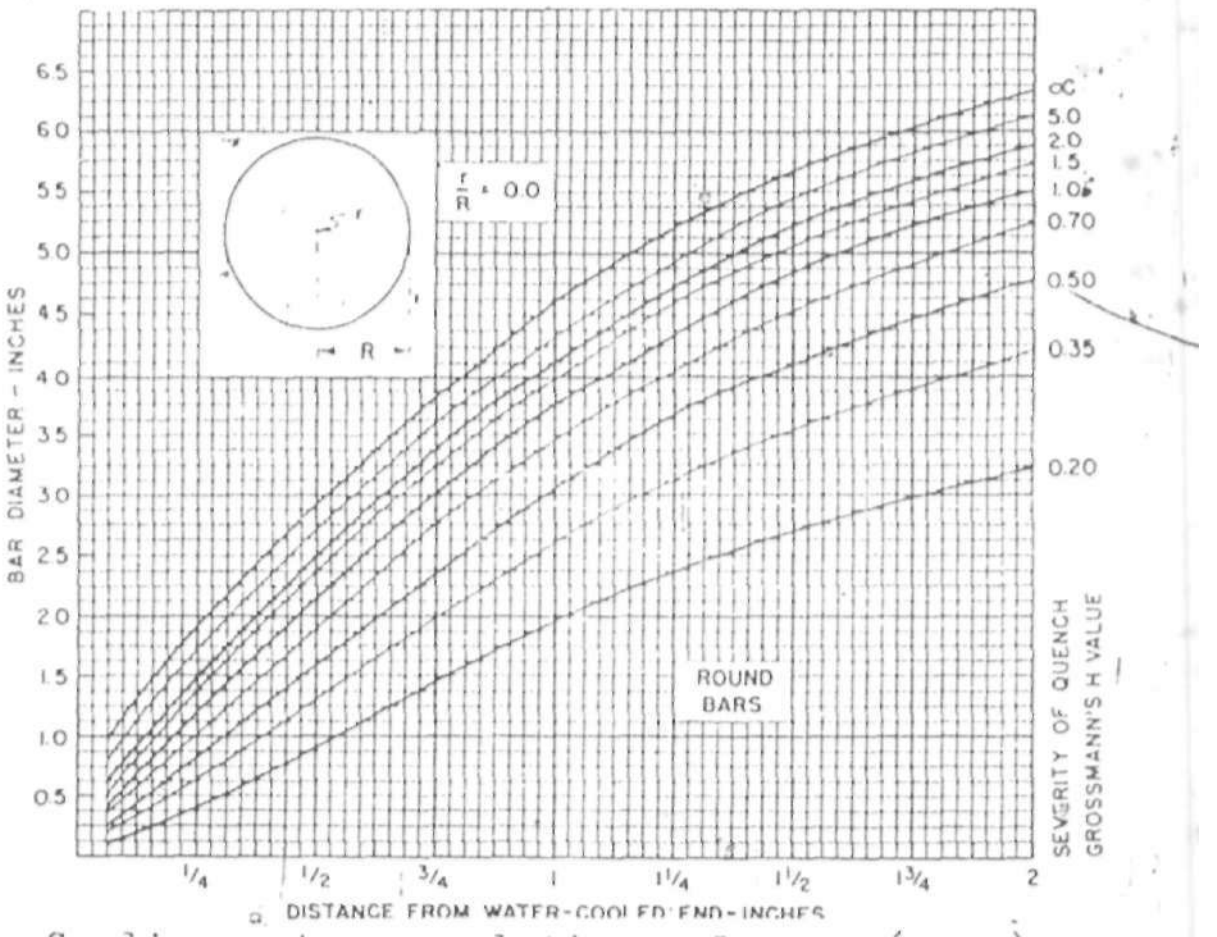
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APPENDIX

Cooling Rate Correlations



Cooling rate correlations; Lamont (1943).