STUDIES

IN   THE

TRANSFORMATIONS   OF   AUSTENITE

by

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C O N T E N T S

INTRODUCTION

The work described in this paper falls into three sections dealing with different though related aspects of the subject.

The first section, dealing with the historical development of the metallography of carbon steels serves as an introduction to the others.   It is not intended as a new contribution to the history of this complex subject but was drawn up in the course of the writer's reading to help him to view the subject as a whole by picking out what seemed to him the most important developments and setting them down in order.   References for this and each of the other sections will be found at the end of the section concerned.

The second section deals with the mechanical properties of austempered steels.   The literature contains very different accounts of the results obtained by this method of heat treatment and it seemed that it might be a profitable field of research.   A series of tests, described in this section, gave results so much inferior to those obtained by more conventional heat treatments that it became apparent that the method was capable of giving either very good or very bad results but that the factors controlling the results were largely unknown.   A large scale investigation would be necessary to achieve any results of real practical importance and facilities were not available for the

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very large number of mechanical tests which would be required.

The final section is devoted to experimental work dealing with the effect of inhomogeneity in austenite on its mode of transformation, principally in steels of about 0.4% carbon. The earlier work in this section, using electrical resistance as a method of following the transformation and microscopical study of isothermally transformed specimens did not produce the results hoped for. The reasons for this failure are discussed. The work at least indicated the possibilities and limitations of the methods. The later work described in this section, principally microscopical examination of heat treated specimens gave more definite results. The effect of inhomogeneity in austenite due to incomplete diffusion of carbon into areas occupied by ferrite in the original structure is examined and it is shown that inhomogeneity due to this cause can, in certain circumstances, play an important part in controlling the response to heat treatment of medium carbon steels but that such inhomogeneity is rapidly removed when the steel is held at the heat treatment temperature. The writer's thanks are due to Professor W. R. D. Jones, D.Sc., University College of South Wales and Monmouth, for carrying out the small scale tensile tests of which the results are given in Table III in this section.

HISTORICAL DEVELOPMENT OF THE METALLOGRAPHY

OF CARBON STEEL.

Although iron was produced and used in considerable quantities in antiquity and cast iron and steel were regularly produced during the middle ages, it was not until about 1800 that the part played by carbon in these materials was fully appreciated and not until almost 1900 that much quantitative information on the effect of carbon was available.

Early writers on metallurgy were chiefly concerned with the extraction of metals from their ores.   One of the earliest printed books on metallurgy was Biringuccio's "Pirotechnica" published in 1540.   In 1556 Agricola produced his magnificiently illustrated "De re Metallica"(1) (one of three books of the same name published before 1570) which contains a vast amount of information on process metallurgy of which a surprisingly large proportion is still applicable.   They do not, however, make any serious attempt to explain the nature of metals and alloys.   The theories of the time, which had some success in chemistry were quite inadequate to explain the phenomena of alloying and heat treatment although a few attempts were made.   Becher and Stahl attempted to explain the hardening of steel in terms of the phlogiston theory while Barba(2) writing in 1640 devotes a chapter of his book to

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a defence of the theory that metals are composed of
the alchemists' sulphur and mercury.   Webster in
1671(3) describes in great detail the "vegetative
growth" of metals and solemnly exhorts his readers
to seek diligently the tree which produces gold.
He does not consider it worth discussing the pos-
sibility of transmutation but goes on to discuss
the mechanism of the process.   It should be noted
that all these writers describe qualitatively a
large number of ores, metals and alloys with an ac-
curacy which is not often equalled at the present
day.

     Reaumur in 1722 attempted to explain the
hardening of steel in terms of the older chemical
theories but in his case his work was supported by
careful experiment and observation.   He suggested
that when the steel is heated "sulphur and salts"
are driven out of the "particles" into the inter-
stices and on quenching they are prevented by the
rapidity of the cooling from re-entering and remain
to cement the particles together.   On tempering they
partially re-enter the particles.   It is possible
to see in this analogies with the present day
theories, substituting carbon for "sulphur and salts",
cementite for "particles" and seeing in his use of
"interstices" a forecast of the modern theory of
solid solubility of carbon in iron, but, of course,
Reaumur's ideas were very much simpler and less
precise /

'precise than this and his advance was not as great as the analogy would suggest. Reaumer's work on grain size, which will be mentioned again, was much more important and did lay a good foundation for later work.

It was not until the work of Bergman in Upsala in 1781 that the importance of carbon in the ferrous alloys was fully realised and the behaviour of wrought iron, cast iron and steel was adequately described in terms of carbon. This was later confirmed by several direct syntheses of steel from iron and carbon, notably that of Pepys and Davy in 1815. Even then the importance of carbon in steel was not accepted by everyone. Dalton, writing in 1827(4) admits that carbon is an essential constituent of cast iron but denies its importance, and, in fact, doubts its existence, in steel. However, in an addendum to the same book he describes an experiment in which he found that steel "formed by Mr. Macintosh's process of cementation with coal gas" when dissolved in sulphuric acid gave off hydrogen containing carburetted hydrogen in an amount corresponding to a carbon content of $\frac{5}{8}\%$ of the steel. He comments, "whether such a quantity can be deemed an essential or an accidental ingredient of steel, may be a subject of consideration". Faraday(5) showed that the residue left when annealed steel was attacked by hydrochloric acid was a carbide though it was not until /

until 1888 that the carbide was shown, by F. Abel,
to have the formula Fe_3C.   It had long been sus-
pected that iron existed in a number of allotropic
forms.   As early as 1600 Gilbert had shown the
existence of the magnetic transformation now known
as the Curie point but it was not until 1869 that
the bearing of allotropy on the practical heattreat-
ment of steel was appreciated.   In that year
several papers were published which mark the be-
ginning of the use of thermal analysis as a means
of investigation, a method which is still one of
the most important in studies of alloy systems.
Tchernoff showed that steel cannot be hardened by
quenching unless it is heated to above a critical
temperature which he denoted by "a", the symbol
still in use (generally printedA).   At the same
time Gore noted that there was a momentary elon-
gation as the steel passed a certain temperature
on cooling and Barrett in 1873 observed thatthis
was accompanied by a rise of temperature, the first
accurate measurement of recalescence though it is
interesting to note that practically a hundred years
earlier J. F. Angerstein mentions the corresponding
absorption of heat (observed visually) at the
critical temperature on heating and states that
this can be used to determine the correct moment
to quench a steel.   The critical temperature ob-
served by these earlier workers was the one marking
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the formation of pearlite and it was, of course,
most marked in high carbon steel.   The upper
critical points are more difficult to observe and
were not discovered until, in 1893, F. Osmond (6)
used the recently invented thermocouple and chrono-
graph to plot cooling curves.   He found two or,
in some cases, three critical points which he called
A_1, A_2, A_3 and he noted the difference in these
points on heating and cooling.   The present day
notation A_c and A_r for temperatures on heating and
cooling respectively was introduced by Osmond.   He
also noted the depression of these points on rapid
cooling or on alloy additions to the steel.   Later
work with thermal analysis is largely an elaboration
of the results obtained by Osmond.   Sauveur(7)
plotted Osmond's points on a diagram and produced
a primitive equilibrium diagram and in 1900 Rooze-
boom(8), interpreting the earlier results in the
light of the phase rule produced an equilibrium
diagram which is reproduced in Fig. 1.   The simil-
arity between this and the diagram accepted at the
present time shows clearly the standard of accuracy
reached by these pioneers in this difficult field.
At this time several papers dealing with the appli-
cation of the phase rule to iron carbon alloys were
published(9).

The existence of δ iron was not proved
until 1914 when it was detected by Ruer and Klesper.

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FIG. 1. IRON-CARBON EQUILIBRIUM DIAGRAM.

The systematic use of the microscope in metallographic work was introduced about the same time as thermal analysis. Isolated instances of its use are recorded, for example Hook's(10) well-known observation of the edge of a razor, and no doubt many others occurred. Reaumur, in 1722, von Fuchs in 1851 and Martens in 1878 all used the microscope to examine fractured surfaces and obtained results of considerable value. Martens also polished and etched specimens and introduced the use of heat tinting as a method of developing structure. The chief credit, however, must be given to H. C. Sorby(11)(12). Sorby not only developed methods of preparing micro-specimens which are essentially the same as those in use at the present time but he evolved, as a result of his observations, a theory to explain the phenomenon of heat treatment which has since been shown to be wrong in some respects but which served as a basis for further work until well into the present century. The idea of polishing and etching metals to reveal their structure was not new. It had been used in 1808 by Alois de Widmanstätten in Vienna to reveal in meteorites, the structure which still goes by his name and which he recognised was due to the crystalline nature of the metal. Sorby examined many samples of steel but he was particularly interested in the "pearly constituent" (later named "pearlite" /

"pearlite" by Howe). He recognised it to be a compound of iron, free from carbon, and a very hard compound of iron and carbon. He believed that at high temperatures, above the critical temperature, iron and carbon combined and that on cooling past the critical temperature the compound broke down into the "pearly constituent" and pure iron. He was unable to see any carbide in quenched steel but was unable to decide whether this was due to the particles being too small to see or to the high temperature compound being fixed by the rapid cooling. He observed the separation into carbide and soft iron which takes place on tempering. Sorby recognised clearly the connection between the critical temperature (only the A_1 change had been recognised at that time), hardening, recalescence and the changes in structure which he had observed.

Sorby's work was soon followed by others and the subject developed rapidly but without any fundamental changes for a considerable time. In 1904 Osmond(13) published his text book of microscopical work in which he described most of the present day laboratory techniques. He also showed that both α and γ iron form cubic crystals and that the slip bands which he described are the 112 and 111 planes respectively. In 1910 the first edition of Deech's Metallography(14) appeared and, in one or other of its editions, this book has been

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a standard textbook ever since and still shows no
sign of being superceded.   Later developments in
microscopical metallography have been largely con-
fined to the development of new etching methods.
The electron microscope has been applied to metal-
lography but, so far, its use has not yet produced
any fundamental advance in the theory of iron -
carbon alloys though Trotter and McLean(15) have
published the first part of an investigation into
the mechanism of tempering of martensite which
seems likely to lead to important results which
would be unobtainable by any other method.

By 1914, as mentioned above, the exis-
tence of α β γ and δ iron had been recognised
and in 1919 A. W. Hull(16) applied the newly
developed method of X-ray crystal analysis to iron
and showed that α iron had a body centred cubic
structure.   From 1921 to 1924, Westgren and, later,
Phragmen (17)(18) published a series of papers in
which they described a systematic investigation of
all the allotropes of iron.   They showed that α ,
β and δ had body centred cubic lattices while γ
was face centred.   Since the α to β change was
not due to any crystallographic change it was con-
sidered that the two forms were not different allo-
tropes /

allotropes and both are now referred to as α iron.
It also seemed about certain, from their work, that
austenite was an interstitial solid solution of car-
bon in iron but it was not possible to show the
position of the carbon atom within the iron lattice.
This was not possible until 1942 when it was ac-
complished by Petch(19).   With the elucidation of
the structure of cementite by Lipson and Petch(20)
in 1941 the problem of the allotropy of iron and
the equilibrium relations between iron and carbon
can be regarded as solved.

The mechanism of the breakdown of austen-
ite is complex and much remains to be done.   The
work of Osmond has already been mentioned: J. E.
Stead in 1914 and H. M. Howe and A. G. Levy(21)
in 1916 continued this work and noted the effect
of transformation temperature (varied by rate of
cooling) on the pearlite structure.   They also
noted the effect of austenising temperature, the
lower the austenising temperature the coarser the
pearlite.   They did not appreciate the importance
of grain size but believed that this effect was due
to inhomogeneity in the austenite.   Their work was
carried further by Portevin and Garvin(22) in 1919
who extended it to include the martensite trans-
formation which had not previously been studied
so intensively as the higher temperature transfor-
mations.   They also gave a quantitative meaning
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to "hardenability" by their work on the quenching
of rounds of varying diameters.   This has been the
subject of intensive study in recent years.   The
rate of reaction in the breakdown of austenite has
been studied by many different methods.   The sub-
ject was first approached by W. R. Chapin in 1922
using the appearance of ferromagnetism as a criterion
of the breakdown.   He found the rate of reaction
just above the martensite point to be very low.   The
use of isothermal transformation, which has proved
so fruitful a method in investigation and, to a much
more limited extent, in practical heat treatment was
developed chiefly by E. S. Davenport and E. C. Bain
(23).   In 1930 they described an isothermal trans-
formation diagram based on the behaviour of unstable
austenite produced by quenching into a bath held at
the temperature concerned below the equilibrium
transformation temperature.   They gave examples
of several of these curves for steels of varying
composition and many more have since been deter-
mined.   Since ~~this time~~ *then* isothermal transformation
has been constantly used as a method of investigat-
ing the breakdown of austenite.   The course of the
reaction can be followed microscopically by quench-
ing samples after varying reaction times so as to
convert any remaining austenite to martensite or,
more conveniently for quantitative work, by means
of the dilatometer.   One point of importance in
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the present investigations arises from Davenport's
and Bain's work.   They drew attention to the con-
stituent (later named Bainite) produced by trans-
formation at temperatures between about 500°C and
200°C, depending on the composition of the steel.
They claimed that this constituent possessed mech-
anical properties, in particular impact strength,
markedly superior to quenched and tempered structures
produced by conventional heat treatment.   The heat
treatment designed to produce bainite was named aus-
tempering.   The mechanical properties of austem-
pered steels are discussed in another section of
this paper.

Davenport and Bain's work has been very
largely confirmed by later investigators except in
regard to the formation of martensite.   Their dia-
gram represents this as taking a short but quite
definite time.   They were not able to measure this
time but assumed that a time factor was involved
though Bain(24) in 1926 had suggested a mechanism
for the formation of martensite which is now gener-
ally accepted and which does not involve any time
factor.   Almost at the same time as Davenport and
Bain's work in America, Robertson(25) in this
country was working on the formation of martensite
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and came to the conclusion that no diffusion took place and therefore time did not enter into the matter. The crystallography of the various constituents involved was studied intensively, in particular the relationships between the orientation of the austenite grain and those of the constituents formed from it. Fink and Campbell(24) showed that martensite had a body centred tetragonal lattice with an axial ratio very close to unity so that the structure resembled that of α iron much more than that of γ . Honda and Mishizama stated, in 1932, that two forms of martensite existed with tetragonal and cubic lattices which they denoted by α and β . The change from α to β takes place on tempering at very low temperatures. This suggestion was opposed by Hägg(26) chiefly on the grounds of volume changes on tempering but Honda's view is now generally accepted. Finally it was shown by Greninger and Troiano(27) in 1940 and by Smith and Mehl(28) in 1942 that martensite is formed by a very complicated process of shear along planes which do not bear any simple relationship to the austenite crystal lattice. Carpenter and Robertson(29) in 1939 suggested that martensite is formed when austenite passes, on cooling, through a range of temperature which depends on the composition of the steel. At high carbon contents this range extends below room temperature so that when a steel is quenched /

quenched down to room temperature it does not pass completely through the martensite range and so transformation of the austenite is not complete and some is left as "retained austenite". The following year it was shown experimentally by Greninger and Troiano(30) that this assumed mechanism is correct and the martensite range is now generally shown on isothermal transformation diagrams instead of the "incubation period" of Davenport and Bain. The rate of cooling through the martensite range is practically immaterial and this fact is utilised, with considerable advantage, in the heat treatment process known as martempering. If cooling is interrupted within the martensite range, Greninger and Troiano found that, provided the temperature is not too low, the austenite which has not already transformed to martensite will remain apparently unchanged for a time and will then begin to transform to martensite and ultimately will be completely transformed. This isothermal transformation of austenite to martensite has been studied intensively in this country by F. C. Thompson and co-workers. Thompson and Stanton(31) in 1945 in a paper covering a much wider field, showed the importance of stresses caused by quenching and by the initial transformation on the rate at which the subsequent transformation proceeds. In a later paper, in 1950, Thompson and Jepson(32) carried this work further /

further and showed that the product of the iso-
thermal transformation is martensite but that the
process is complicated by the tempering of the mar-
tensite formed by quenching and, later, of that
formed isothermally.   They conclude that in both
cases martensite formation is the result of stresses
producing movement on some definite crystallographic
plane in the austenite though it is not certain
that the plane concerned is the same in both cases.

The decomposition of austenite at higher
temperatures to form pearlite has been studied in
great detail.   It is now known to be a process of
nucleation and growth.   The process has been treat-
ed mathematically by, amongst others, Johnson and
Mehl(33) in 1939 and Hull, Colton and Mehl(34) in
1942 and it has proved possible to determine rates
of nucleation and growth in many cases.   The
nucleus, in most cases, is a platelet of cementite
formed on a grain boundary.   At high temperatures
few nuclei are produced and growth is rapid, re-
sulting in a structure showing large pearlite areas,
often much larger than the austenite grain in which
they originated.   At lower temperatures more nuc-
lei are produced and growth is slower so that the
final structure consists of small, roughly semi-
circular patches of pearlite extending from the
grain boundary at which they were nucleated towards
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the centre of the grain.   The actual rate of for-
mation of pearlite under these conditions has been
found to agree in a most satisfactory manner with
the mathematical treatment and according to Mehl(35)
"no mystery on this aspect of the formation of pearl-
ite remains".   Heterogeneity in the austenite serves
to produce a much larger number of nuclei, within the
grain as well as at the boundary, and the resulting
effect on the rate of transformation has been used
as a test for the attainment of homogeneity(36).
This work is further discussed in the third section
of the present paper.

The formation of bainite from austenite
has not received as much attention as the trans-
formations at higher and lower temperatures.   It
seems, however, that it also is a process of nucle-
ation and growth.   In this case the process is
nucleated by ferrite which may be precipitated
first in an unstable form containing carbon in
supersaturated solution, though this has only
been definitely observed in alloy steels(37).

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MECHANICAL PROPERTIES OF AUSTEMPERED STEELS

By austempering is meant the rapid cooling of a steel from a temperature above the AC_3 to a lower temperature at which it is held till the austenite has transformed completely to bainite, the temperature depending on the composition of the steel and the mechanical properties required. The process was the subject of an American patent and is claimed to give mechanical properties superior to those obtainable by the conventional quenching and tempering treatment. In particular it is claimed that the impact resistance corresponding to a particular hardness obtained by austempering is much greater than that found when the steel is quenched and tempered to the same hardness. The first comprehensive investigation of the method after its introduction in 1930 by Davenport and Bain (1) was published in 1934 by Davenport, Roff and Bain (2). They investigated a series of plain carbon steels ranging from 0.50% to 1.17%. Table I shows the results obtained on one of these steels. No other results are given but the results on this steel are said to be "quite typical" of all the comparisons. It will be noted that while the steel shows the same hardness in the two cases (which is arranged by selecting a suitable tempering temperature) the properties of the austempered steel are markedly /

markedly superior, particularly in regard to ductility and impact resistance. The hardness can be varied over a fairly wide range and it was found that for all hardness values austempering showed superior impact resistance but that the superiority was not so marked at very high or very low hardness values.

TABLE I

Mechanical properties after austempering and after oil quenching and tempering

Composition   C. 0·74%   Mn. 0·37%   Si 0·145%

Heat Treatment

I	II
Austenise 5 min. at 790°C	Austenise 5 min. at 790°C
Quench into lead alloy bath at 305°C	Oil quench
Hold 15 min.	Temper 30 min. at 315°C
Water quench	Water quench

Mechanical Properties		I	II
Hardness.	Rockwell C	50.4	50.2
Maximum stress.	lbs./sq.in.	282,700	246,700
Yield point	lbs./sq.in.	151,300	121,700
Elongation	%	1.9	0.3
Reduction of area	%	34.5	0.7
Impact strength.	ft.lbs. on 0.180" dia. un-notched bar	35.3	2.9

The authors ascribe the improved properties to the absence in the austempered specimens of micro-cracks which, /

which, they claim, are almost invariably found in quenched and tempered specimens after etching. The amount of cracking varies directly with the austenitic grain size of the steel before quenching and this accounts for the well known reduced impact values found in coarse grained steel. The impact resistance of austempered specimens, free from micro-cracks, shows only slight variation with grain size.

Since Davenport, Roff and Bain's paper a great deal of work has been carried out on isothermal transformations of austenite but much the greater part of this has been devoted to the construction of S-curves and to elucidating the mechanism of the transformation in alloy steels of widely varying composition. Comparatively few data have been published of the mechanical properties obtainable by this method in the heat-treatment shop. The investigations which have been published show widely varying results and, on the whole, do not bear out the comprehensive claims made by the earlier workers. Hughes and Dowdell (3) confirmed the marked improvement in toughness produced by austempering but Payson and Hodapp (4) a few years later found that in a large number of the low alloy steels covered by the SAE specifications austempering was not advantageous.

Legge(5) reported that an eutectoid steel gave markedly improved results on austempering in thicknesses. /

thicknesses of, at least, up to 0.18". Griffiths, Pfeil and Allen(6), working with alloy steels, found that bainite can be produced by continuous cooling in many cases by a transformation intermediate between the pearlite and martensite points. In these cases the steels were generally found to possess poor ductility, elastic properties and impact strength. By austempering steels to produce pure bainite they were able to show that these properties were not due to the presence of martensite but were inherent in the bainite itself. Almost without exception their austempered specimens were markedly inferior to quenched and tempered specimens.

Gensamer, Pearsall and Smith(7) investigated the mechanical properties of a 0.78% carbon steel quenched from above Ac_3 temperature to various lower temperatures in an attempt to correlate these properties with the interlamellar spacing of the structures produced. They did not, unfortunately, include impact tests. Their results are not strictly comparable with Davenport's but they seem to indicate that the structures found were considerably softer and more ductile than Davenport's. They also suggest that ductility may be a minimum when the transformation takes place at about 500°C and that lower temperatures may produce increased ductility as well as increased hardness. This is unexpected as it involves a very abrupt change in ductility /

ductility when the upper limit of the martensite
range is reached and brittle martensite is produced.
Elmendorf(8) reported that a mixed structure of
bainite and martensite gave good results in several
cases.

Experimental.

A comparison was attempted between four
steels of different types in the austempered and
quenched and tempered states.   Since improved im-
pact resistance is the main claim made for aus-
tempering, only the impact strength was measured.

The four steels used had the following
compositions.

TABLE II

	ED	EE	EF	EG
Carbon	0.26	0.39	0.26	0.31
Silicon	0.24	0.21	0.16	0.24
Manganese	1.44	0.52	0.51	1.44
Sulphur	0.010	0.013	0.016	0.021
Phosphorus	0.023	0.026	0.025	0.020
Nickel	0.01	0.01	3.54	-
Chromium	0.05	0.03	1.33	-
Molybdenum	0.36	-	-	-

Reference to published isothermal transformation
diagrams(9) shows that complete transformation takes
place at 400°C in 5 minutes for steels ED. EE and
EG and 15 minutes for steel EF which is rendered
more /

more sluggish by the alloying elements present
(isothermal transformation diagrams published since
this section of the work was completed(10) show a
period of about 1 hour for complete transformation
of a 3% Ni 1% Cr steel at 400°C). The heat treat-
ment was carried out in an electric furnace mounted
so that the specimens could be lowered from this
furnace into a gas heated lead bath placed beneath
it. The temperature of the electric furnace was
controlled by a Cambridge potentiometer type con-
troller in conjunction with a Pt - Pt Rh termo-
couple. The lead bath was controlled manually,
the temperature being measured by a copper-constan-
ton thermocouple in a silica sheath.

Two specimens of each steel were rough
machined in the form of Izod test pieces. One
specimen of each steel was held for 30 mins. at
860°C and then quenched into the lead bath at
400°C where it was held until transformation was
complete. The specimens were then quenched in
water. The water quench is not an essential part
of the austempering treatment but was adopted in
this case partly as a matter of convenience in hand-
ling the specimens and partly to avoid the possible
occurrence of temper brittleness in the nickel-
chromium steel. Temper brittleness has never
been reported after austempering but as the
mechanism of its production is still unknown it
seemed /

seemed better to take no risks in this case.   The
hardness of these specimens was measured.   A number
of small specimens of each steel was then water
quenched and tempered at progressively increasing
temperatures until the hardness was reduced to
that of the austempered specimens.   Izod test
piece blanks were then water quenched and tempered
in a muffle furnace for 30 mins. at the appropriate
temperature, followed by a water quench.   The tem-
pering temperatures and the hardnesses produced by
austempering and by water quenching and tempering
are shown in Table III.   It should be noted that
in no case is the austempered specimen harder than
the corresponding water quenched and tempered one.

TABLE   III

| Steel | Tempering temperature | Diamond pyramid hardness | |
		Austempered	Quenched & tempered
ED	650°C	252	257
EE	650°C	209	229
EF	525°C	358	358
EG	650°C	218	227

The test pieces were then machined to
their final size with three notches in each and
impact tests carried out on a standard Izod machine.
The results of these tests are shown in Table IV.
Micro-examination of the broken test pieces showed
little or no martensite indicating that transfor-
mation /

transformation had been practically complete.

TABLE IV

	Austempered at 400°C		Quenched & tempered	
	Impact strength ft. lbs.	Average ft. lbs.	Impact strenth ft.lbs.	Average ft.lbs.
ED	6, 5, 6	6	103, 105, 103	104
EE	72, 77, 83	77	107, 108, 111	109
EF	13, 14, 13	13	52, 53, 54	53
EG	5, 8, 21	12	93, 91, 95	93

In all cases the austempered specimens are markedly
inferior to those which have been water quenched and
tempered.   In the case of steel (EE) which is a
straight carbon steel austempering gives a satis-
factory toughness though much less than that ob-
tainable by quenching and tempering.   The three
alloy steels, however, are very brittle when aus-
tempered though the conventional heat treatment gives
a satisfactory degree of toughness.   These results
are completely contradictory of the early published
claims for austempering.   It is clear that austem-
pering is not advantageous in all cases.   The number
of steels tested was far too small for any general
conclusion to be drawn but the results suggest that
while the impact strength of bainite produced in a
straight carbon steel may be satisfactory, it may be
dangerously low in the case of alloy steels.   In
order to test the possibilities of this process
thoroughly /

thoroughly a very long series of experiments would be necessary. A considerable number of variables, besides composition, would have to be takin into account.

1. Time required for complete transformation at any given temperature. This depends principally on the composition of the steel but also on the other factors and would have to be determined in each case.

2. Transformation temperature. The mechanical properties depend on the temperature at which the steel transforms. It has been stated that they depend more on this than on any other factor. In general, the lower the transformation temperature, the harder, stronger and less ductile the product but it is possible that one particular temperature may give an optimum combination of properties.

3. Austenitic grain size. The time required for transformation depends on the austenitic grain size. This dependence is now well understood but the effect of austenitic grain size on the final mechanical properties is largely unknown.

4. Previous history. Thompson and Stanton(9) showed that the previous treatment of the steel /

steel may have a profound effect on its
response to austempering.   This factor
could no doubt be eliminated by a pro-
longed cook at a high temperature to give
a homogeneous austenite followed by a suit-
able treatment to give some standard initial
structure before the actual austempering
operation.

5. Austenising time.   The extent to which car-
bides enter into solution in the austenite
and the degree of homogeneity achieved in
the austenite itself depend on the tempera-
ture and time of austenising.   Both factors
control the time required for transformation
and probably also the final properties.

6. Size of specimen.   The whole operation of
austempering depends on the rate of cool-
ing from the austenising temperature to
the transformation temperature being suf-
ficiently rapid to prevent the formation
of pearlite of any type.   The larger the
specimen the more difficult it will be to
achieve this rate of cooling.   There is,
therefore, an upper limit to the size of
specimen which can be treated.   The limit-
ing size will depend on the composition of
the steel and also on the nature of the
quenching medium.   With plain carbon
steels /

steels quenched into a lead or alloy bath,
the upper limit of size seems to be of the
order of $\frac{1}{4}$ inch diameter.   With alloy steels
larger sizes should be possible in many cases.
In the experimental work described above,
it was realised that the size of specimen
was about the upper limit of size which
could be treated successfully but this was
unavoidable as it was not possible to carry
out impact tests on smaller specimens than
the standard Izod test piece.   The large
specimen used together with the high trans-
formation temperature, 400°C, made it likely
that some pearlite would be produced but
careful micro-examination of the broken
test pieces did not show any constituent
which could be identified as a form of
pearlite.

It was intended to carry out a series of
tests to investigate the possibilities of austem-
pering as a practical means of heat treatment but
this was found to be impracticable and the idea
had to be abandoned.

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The Transformation of Austenite with Particular Reference to the Effect of Inhomogeneity.

The breakdown of austenite, either on continuous cooling or isothermally, consists essentially of the change from the face centred cubic γ solid solution to the body centred cubic α form, accompanied by more or less separation of the dissolved carbon as cementite. The way in which this transformation takes place is influenced by many different factors and its study is the basis of the whole theory of heat treatment of steel. So far as the practical application of heat treatment is concerned the most important difference between γ and α iron is their differing ability to dissolve carbon but in studying the transformation other properties are frequently used to follow the course of the change. In the earlier work on this subject and at the present time when the transformation is taking place during continuous heating or cooling, the standard methods of following the progress of the change are thermal analysis and dilatometry. Thermal analysis is not applicable to isothermal transformation and Davenport and Bain (1) whose paper in 1930 drew general attention to the possibilities of this method of heat treatment, used a microscopical method of attack. Specimens of the steel being studied were quenched from a high temperature at which they were wholly austenitic into a bath of low melting point alloy held at the temperature /

temperature at which the steel was to be transformed. After varying intervals of time the specimens were withdrawn from the bath and quenched in water. The specimens were then microscopically examined and the extent of transformation in the bath could be measured since the untransformed austenite was converted into martensite by the water quench and could be clearly distinguished from the product of transformation at the higher temperature. This method of approach is laborious and time consuming but it has the advantage that the mode of formation of the transformation product can be studied. Where the rate and extent of the transformation is required rather than the details of the mechanism, the quenching dilatometer is now the standard instrument. Several forms have been described(2). Methods based on the differing magnetic and electrical properties of α and γ iron have also been used successfully(3).

Electrical Resistivity.

The electrical resistance of a metal specimen can readily be measured with a fairly high degree of accuracy. Since there is a considerable difference between the specific resistance of austenite with carbon in solution and ferrite it seemed that this could be utilised conveniently as a means of studying the transformation. In order to have a readily measurable resistance the specimen would require /

require to be in the form of a wire some inches in
length.   Preliminary tests were carried out to see
whether the equipment available could be used to ad-
vantage.   Continuous heating and cooling were used
as being simpler than isothermal transformation and
yet providing an adequate test of the equipment.   The
test piece and layout are shown in Fig. 1.   The
material used was steel containing o.55% C, 0.62%
Mn. in the form of 10 S.W.G. wire.   The test piece
was heated and cooled in a muffle furnace the tempera-
ture of which was measured by means of a chrome-alumel
thermocouple with its junction in the centre of the
coiled test piece.   The test piece was small enough
to be entirely within the uniform temperature zone
of the furnace.   The current through the test piece,
from two or three accumulator cells, was controlled
by a rheostat and, in the earlier experiments, was
adjusted to a fixed value whenever any variation was
observed.   In later experiments it was found that
more consistent results could be obtained by compar-
ing the potential drop across the test piece with
that across a similar specimen kept at a constant
temperature in a vacuum flask.   The potential drop
was measured by means of a Tinsley constant resis-
tance deflection potentiometer.   Typical results
are shown in Fig. 2.   The beginning and end of the
transformation /

FIG. 1. ELECTRICAL RESISTANCE.

TEST PIECE   STANDARD

POTENTIOMETER

A

TEST PIECE

HALF SIZE

THERMOCOUPLE

POTENTIAL LEADS

CURRENT LEADS
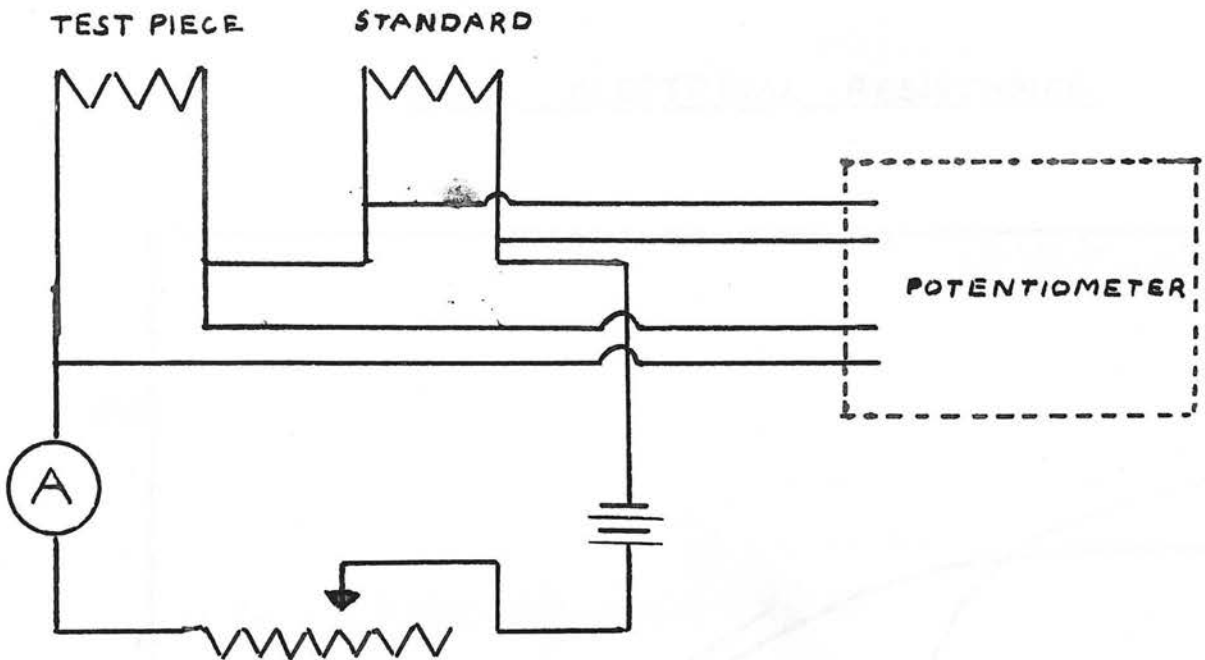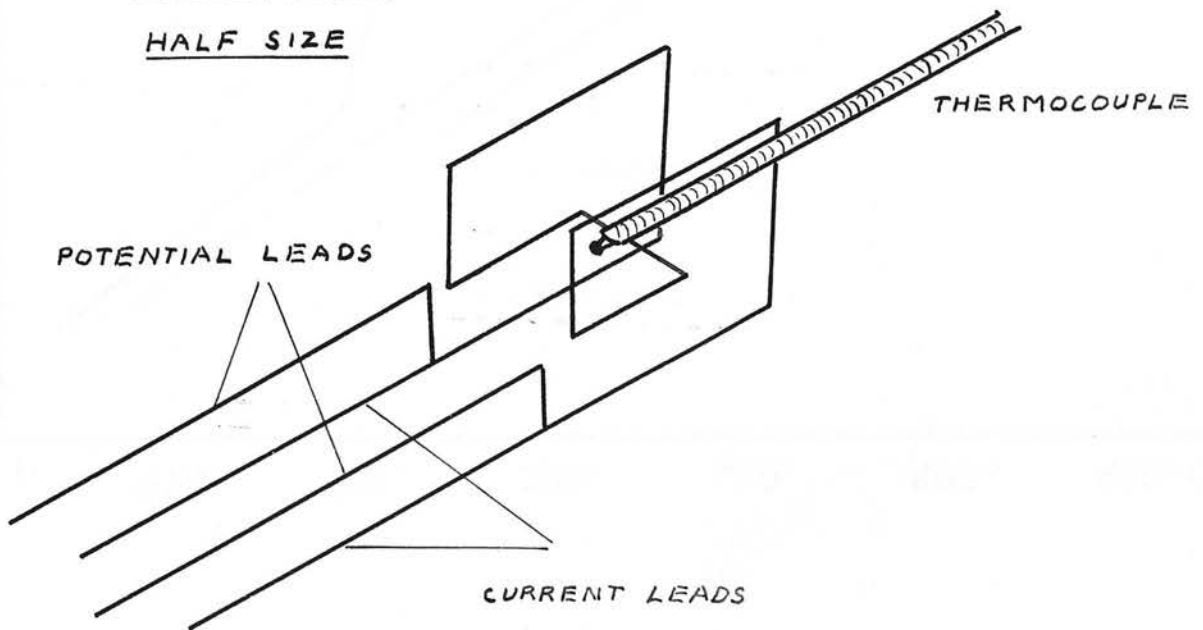
FIG. 2.   ELECTRICAL   RESISTANCE.
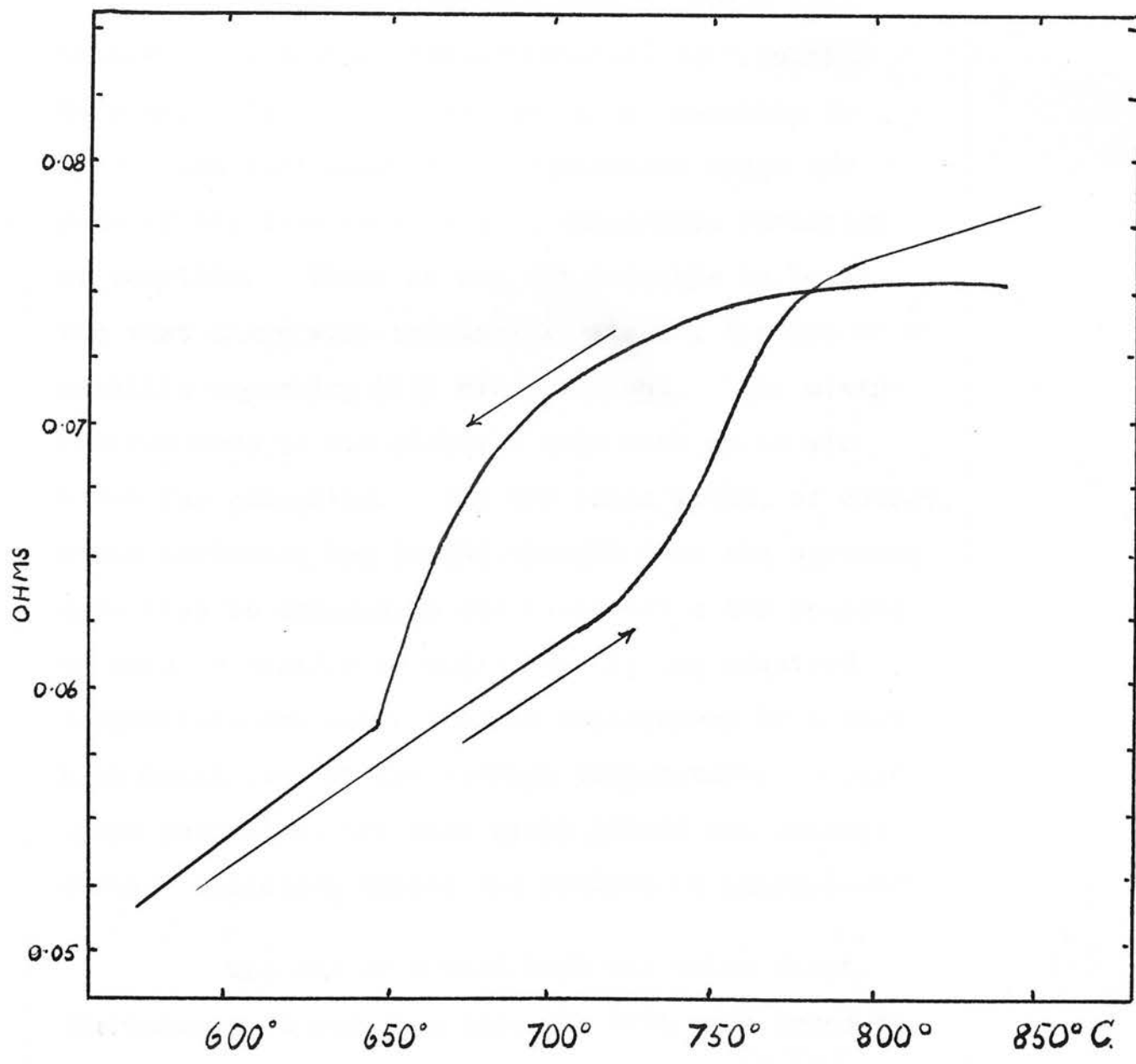
transformation are well marked and an estimate of the
rate of transformation could be made by comparing the
slope of the curve with the slope before and after
transformation.

When an attempt was made to apply this
method to isothermal transformation, difficulties
were encountered.   It was not found possible to
quench the test piece to a temperature below the
nose of the S-curve without considerable formation
of pearlite.   Since it was not possible to cover
the test piece with insulating material the use of a
metallic quenching bath was ruled out.   The alter-
natives were to use either a salt bath or an air
blast for quenching.   The air blast would, of course,
cause oxidation but it was thought that the specimen
need only be exposed to the blast for a few seconds
to cool it rapidly to approximately the required
temperature and could then be transferred to a salt
bath maintained at the correct temperature.   Under
those conditions the test piece should not undergo
further oxidation during the resistance measurements.

The use of a salt bath was tried first.
Microspecimens quenched into the bath were found to
be largely transformed to pearlite during cooling
and it proved impossible to obtain purely isothermal
reaction.   With more slowly reacting alloy steels
the salt bath would be adequate for quenching but
for /

for plain carbon steels it is too slow.   Attempts to
use an air blast were also unsuccessful.   The blast
was supplied by a small electric air compressor and
was passed through a coil of three turns of iron pipe
heated in a large gas muffle furnace.   The tempera-
ture of the blast could be raised in this way to about
300°C but it was again found impossible to obtain a
satisfactory quench.   These experiments were, there-
fore, discontinued.

Pearlite Formation.

The formation of pearlite from austenite
has been studied in great detail.   Johnson and Mehl
(6) in 1939 and Hull, Colton and Mehl(4) in 1942 des-
cribed the kinetics of the process.   It is one of
nucleation and growth in all cases but the varying
rates of nucleation and growth with temperature re-
sult in two apparently widely differing reactions.
Nucleation takes place in all cases at the austenite
grain boundary.   At high temperatures, approaching
the Ae_1 temperature, nuclei are produced only in
small numbers while the growth of the resulting
pearlite nodules is rapid.   The nodules grow into
approximate spheres until growth in any one direction
is prevented by impingement on another nodule.   Growth
of the nodules is not interrupted by austenite grain
boundaries and the nodules may grow until they are

much /

much larger than the austenitic grain size.   In
these circumstances nucleation is approximately
random though actually confined to the grain boundary.
 It was found possible to derive an equation

$$f(t) = 1 - e^{-\frac{\pi}{3} N_v G^3 t^4}$$

connecting the fraction of the austenite transformed,
$f(t)$, at time t with N_v, the nuclei formed per cubic
millimetre per second and G, the rate of radial growth
in millimetres per second.   This equation gives a
curve which fits closely the isothermal reaction curve
at high temperatures.   At lower temperatures, ap-
proaching the knee of the S-curve, nucleation is more
rapid compared with growth and nucleation is no longer
approximately random.   Impingement of the nodules
takes place much earlier and their growth is largely
confined to the direction of the centre of the aus-
tenite grain.   Very soon the grain is completely
outlined by nodules in various stages of growth.
Under these conditions the mathematical analysis prov-
ed much more difficult.   The final result was ex-
pressed in the form of a set of master curves which
plot the fraction transformed $F(z)$ against $\frac{G}{\alpha}$ x t for
various values of a shape factor $\lambda = \frac{\alpha^3 N_s}{G}$ where α

is the radius of the austenite grain and N_s the number
of nuclei formed per second per square millimetre of
grain surface.   It was pointed out in these two
papers that the presence of undissolved carbide or

carbon /

carbon concentration gradients could act as nuclei
which were not confined to grain boundaries and so
could make the nucleation to some extent independent
of grain size.   This was further developed in a later
paper by Roberts and Mehl(5) and made into a criterion
of the practical homogeneity of austenite.   Assuming
that a steel was reacted at a temperature near to the
Ae_1 temperature and that the rate of nucleation was
proportional to the grain boundary area, they showed
that Johnson and Mehl's equation for random nucle-
ation, given above, implied a linear relationship be-
tween the austenitic fracture grain size and the
logarithm of the time of half reaction and that this
relationship had a constant slope, independent of the
true reaction rate of the austenite.   If the relation-
ship between austenitic grain size and time of half
reaction was found, for a constant reaction tempera-
ture, the presence of inhomogeneity in the austenite
would be shown by a departure from this constancy of
slope.   They showed experimentally that after an aus-
tenising time which depended on the composition of the
steel and on the austenising temperature this relation-
ship held while for shorter austensing times it did
not.   This critical austenising time was taken as
the time required to reach practical homogeneity in
the austenite.

The above work applied strictly only to
eutectoid and hypereutectoid steels.   In the case
of /

of hypo-eutectoid steels the situation is complicated
by the separation of proeutectoid ferrite.   It seemed
probable, however, that differing degrees of austenite
homogeneity should result in changes in the course of
isothermal reaction which would be detectable experi-
mentally even though they were not amenable to mathe-
matical analysis and a series of experiments was car-
ried out with a hypoeutectoid steel and with one which
was practically eutectoid.

Heat Treatment Equipment.

The equipment used for heat treatment is
shown in Fig. 3.   The high temperature furnace con-
sisted of a tubular furnace mounted vertically.   An
alundum combustion tube was supported centrally in
this by means of two asbestos boards, one at each
end of the furnace.   At the upper end of the alun-
dum tube a short length of copper tube, large enough
to fit loosely over the outside of the alundum tube,
was attached.   A side tube admitted nitrogen while
the step at the top of the alundum tube served to
support a cross piece of thick steel wire from which
the specimens were hung.   The top of the copper tube
was closed with a bung through which a chromel-alumel
thermocouple passed.   A disc of asbestos sheet on
the foot of the bung protected it sufficiently from
the heat.   The lower end of the alundum tube was
closed by a bung through which the exit tube for the
nitrogen /

FIG.3. HEAT TREATMENT EQUIPMENT.

← THERMOCOUPLE

← RUBBER BUNG

← NITROGEN IN

COPPER FITTING →

← 1½" INT. DIA. ALUNDUM TUBE

ASBESTOS SHEET

WIRE SUPPORT

SPECIMEN TIED TO SHORT WIRE.
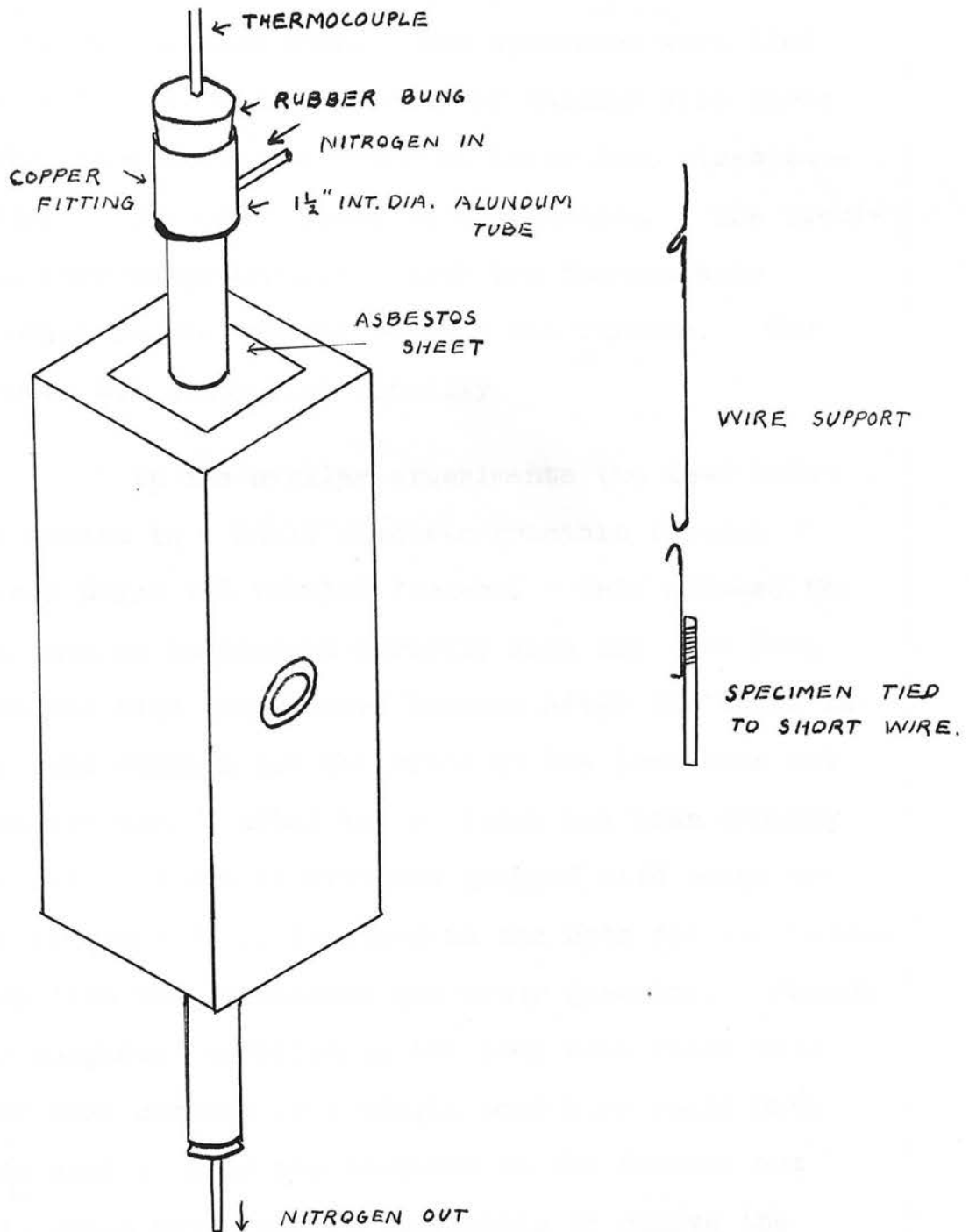
↓ NITROGEN OUT

nitrogen passed.   This bung was similarly protected
by a disc of asbestos board.   The outgoing nitrogen
passed through a rubber tube to a double bubbler to
indicate the rate of flow and avoid sucking back when
the furnace cooled down.   The specimens were tied
with thin iron wire to pieces of thicker wire three
inches long which were hung by hooks from wires sup-
ported by the cross piece in the furnace.   The speci-
mens were supported level with the thermocouple
junction in the central zone of the furnace.   The
furnace was controlled manually.

In the earlier experiments the lead bath
was heated in a small electric crucible furnace
placed below the tubular furnace.   This allowed the
specimen to be dropped directly into the lead bath
from the high temperature furnace after the bungs in
the tube furnace and the cover on the lead bath had
been removed.   After the specimen had been dropped
the short length of wire was gripped with tongs and
the specimen fully immersed in the bath for the neces-
sary time then withdrawn and water quenched.   Prompt
and complete immersion in the lead bath would have
been more certain if a single long wire could have
been used to hang the specimen in the furnace but
this would have made it impossible to remove the
specimen readily from the lead bath.   This lead
bath was found to be too small.   With unusually
large specimens or a group of small specimens
quenched /

quenched simultaneously the temperature in the bath
rose appreciably and it was found impracticable to ad-
just the initial temperature so that the final tempera-
ture would be correct.   A large gas heated lead bath
was, therefore, used.   This was a simple iron pot
surrounded by bricks and heated by a gas ring.   The
heat given off above the bath was too great to allow
it to be used below the electric furnace.   Instead,
it was placed at the side and the specimens were mount-
ed on a single long wire by which they were withdrawn
from the top of the electric furnace.   This arrange-
ment proved easier to use and the delay in quenching
was very little more than with the electrically heated
lead bath.   There was no measurable rise in tempera-
ture when specimens were quenched.   The temperature
was measured by means of a chromel-alumel thermocouple.

Materials and Preliminary Experiments.

The steels used in this series of experiments
had the analyses shown in Table I.

TABLE I

	C	Mn	Si	S	P	Ni
A	0.41%	0.58%	-	-	-	-
B	0.45	0.68	0.13	0.043	0.020	0.10
C	0.74	0.52	-	-	-	-

Steels A and B were in the form of 5 S.W.G.
(0.212" diam.) rod while C was 0.31" diameter rod.
Steel /

Steel A was used in the earliest experiments but was later replaced by B as the available supply of A was exhausted.

The critical temperatures of these steels were determined by means of a Gale dilatometer. Typical heating and cooling curves are shown in Fig. 4. The results obtained are given in Table II.

TABLE II

	Critical Temperature °C			
Steel	AC_1	AC_3	Ar_1	Ar_3
A	710	795	655	728
B	715	785	655	725
C	718	745	670	710

The time required for a specimen to reach the critical temperatures after insertion in the heat treatment furnace was estimated. A dilatometer test piece of steel A was used for this purpose. The thermocouple junction was inserted in the hole of the test piece and the opening packed with shreds of asbestos. The test piece was then put into the furnace and temperature readings taken every fifteen seconds. The results are shown on the graph, Fig. 5. The lower critical temperature is passed within $1\frac{3}{4}$ minutes and the upper critical temperature is reached about a quarter of a minute later. These figures, of course, apply only to specimens of this size, 0.21" diameter.

Isothermal /

FIG. 4. DILATOMETER TEST
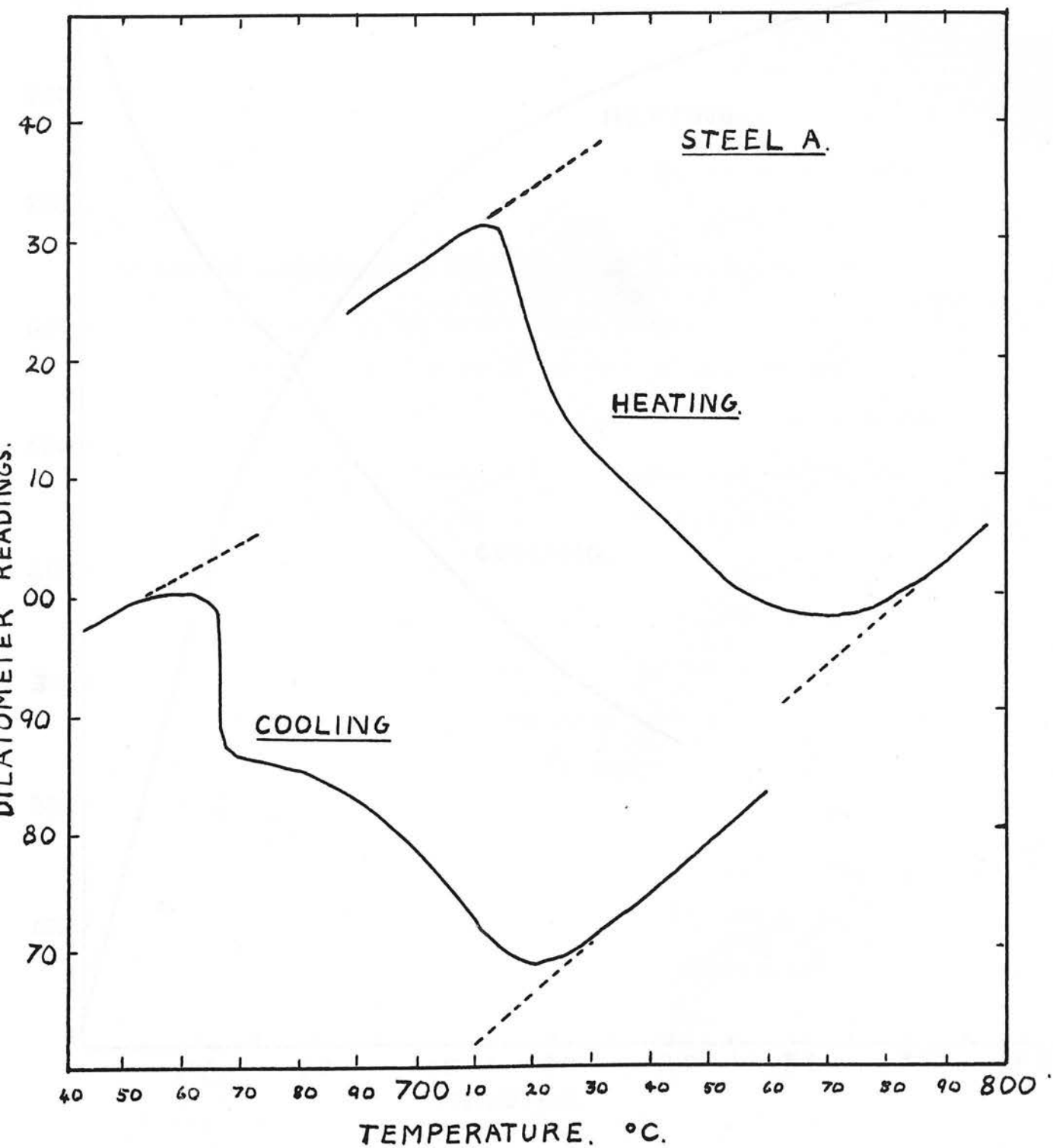
FIG. 5. HEATING AND COOLING CURVES OF 5 S.W.G ROD.
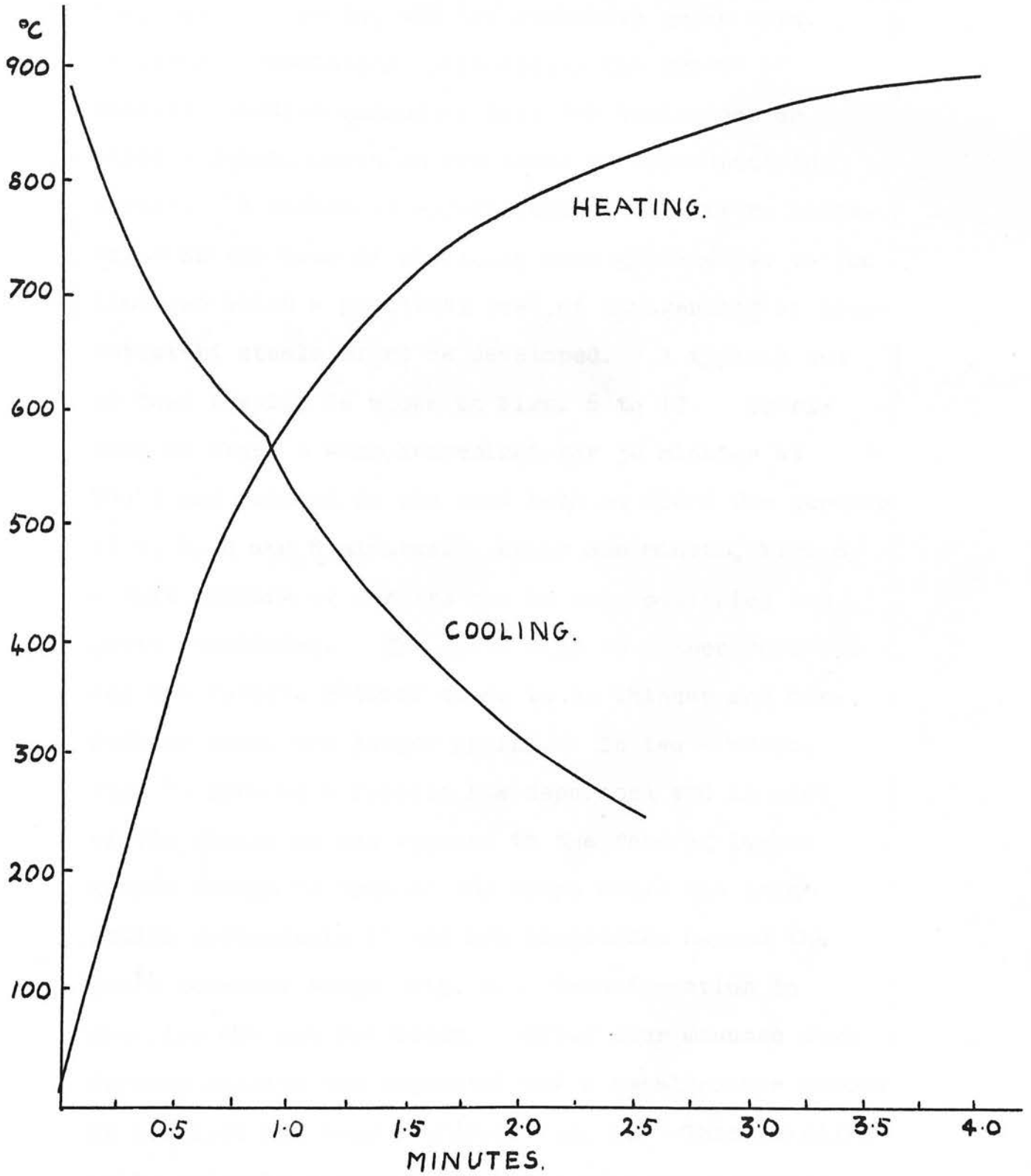
Isothermal Transformations.

Roberts and Mehl's(5) method of detecting
austenitic inhomogeneity depends on the relationship
between the time of half transformation at a tempera-
ture near to the Ae_1 and the austenite grain size.
It assumes conditions, principally the growth of
pearlite nodules unimpeded save for impingence on
other nodules, which do not exist in hypoeutectoid
steels.   A series of experiments was, however, under-
taken in the hope of obtaining some guidance as to the
lines on which a practical test of homogeneity in hypo-
eutectoid steels might be developed.   A typical set
of test results is shown in Figs. 6 to 12.   Speci-
mens of steel A were austenised for 30 minutes at
900°C and reacted in the lead bath at 680°C for periods
of 1, 2, 4 and 8 minutes.   After one minute, Fig. 6,
a thin network of ferrite can be seen outlining the
grain boundaries.   The grain size is rather variable
and the ferrite network tends to be thinner and more
regular round the larger grains.   In two minutes,
Fig. 7, much more ferrite has separated and in most
of the grains it now appears in the form of larger
masses though in some of the areas where the large
grains predominate it has not progressed beyond the
grain boundary stage, Fig. 8.   Transformation to
pearlite has not yet begun.   After four minutes some
further ferrite has separated and a considerable amount
of pearlite has been produced, Fig. 9.   This pearlite
is irregularly distributed.   Its area was measured

by /
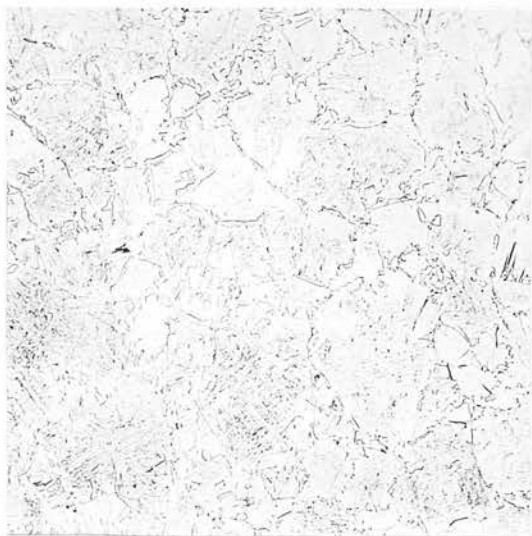
Fig. 6.   Steel A.   Austenised 900°C
     Reacted 1 min. at 680°C x 100.
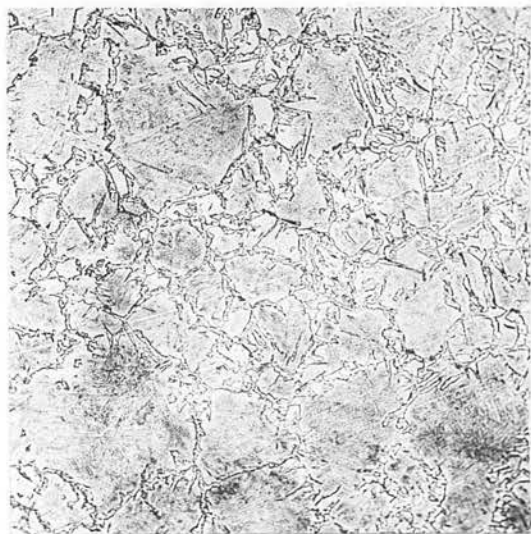
Fig. 7.   Steel A.   Austenised 900°C
     Reacted 2 min. at 680°C x 100.
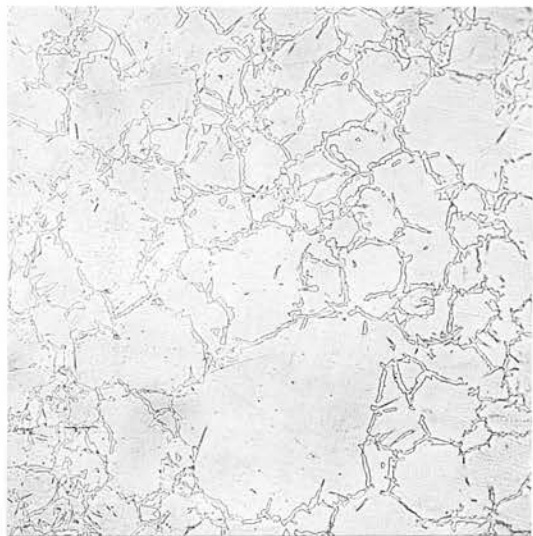
Fig. 8.   Steel A.   Austenised 900°C
     Reacted 2 min. at 680°C x 100.
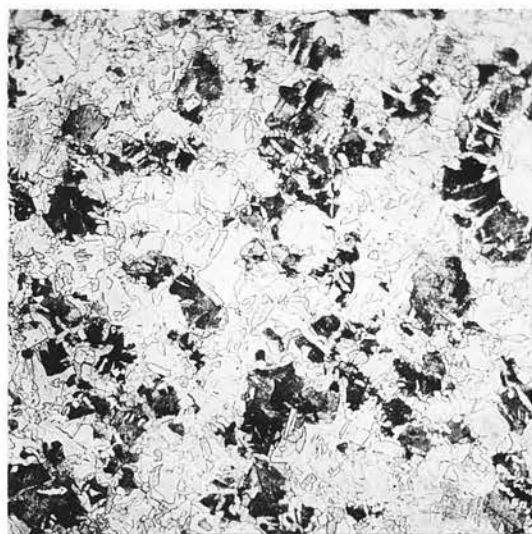
Fig. 9.   Steel A.   Austenised 900°C
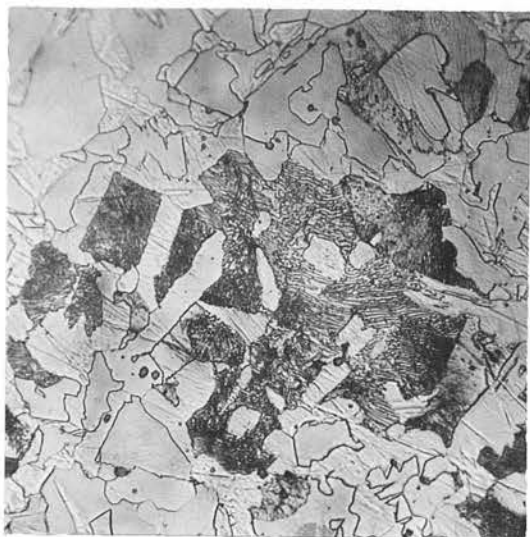   Reacted 4 min. at 680°C x 100.

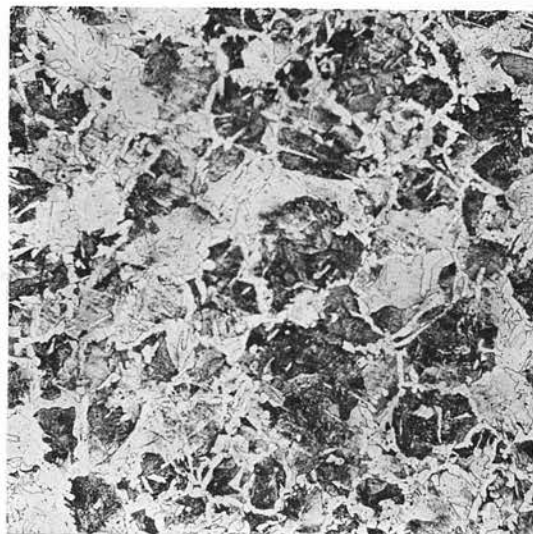Fig. 10. Steel A. Austenised 900°C
Reacted 4 min. at 680°C x 500

Fig. 11. Steel A. Austenised 900°C
Reacted 8 min. at 680°C x 100.

Fig. 12. Steel A. Austenised 900°C
Reacted 8 min. at 680°C x 500.

Fig. 13. Steel A. Austenised 900°C
Reacted 2 min. at 680°C x 100.

by placing the negative over a sheet of thin graph
paper on a glass tracing desk illuminated from below
and counting the squares visible through the different
parts of the negative.   The structure consists of
40% to 45% ferrite, 15% pearlite and the remainder
martensite.   Fig. 10 shows the structure at a higher
magnification.   The laminations of the pearlite in
many cases are unusually straight and run across the
shorter axis of the area in which they are produced.
Lengthwise growth of the cementite platelets is very
quickly interrupted by areas of ferrite and growth of
the pearlite proceeds by nucleation of fresh cementite
platelets.   After eight minutes, as Fig. 11 shows,
transformation was substantially complete.   Fig. 12,
the same specimen, shows the presence of a very small
amount of martensite indicating that transformation
was just being completed at that time.   In this series,
then, formation of pro-eutectoid ferrite began before
the lapse of one minute, between two and four minutes
formation of pearlite began and the transformation was
completed eight minutes after the quench.

Although the results shown above are typical
of the structures produced, the times taken for each
stage were found to vary considerably even when the
experiment was repeated under conditions as nearly
the same as possible.   Fig. 13 shows the same steel
austenised and reacted at the same temperature as in
the /

the series above for two minutes and in this case
transformation has proceeded much further than is
shown in Fig. 9, approximately 40% on the average be-
ing transformed to pearlite though the actual percent-
age varies considerably from one part of the specimen
to another.   The most probable cause of this failure
to obtain consistent results is the difficulty of
maintaining a steady temperature in the small, gas
heated lead bath employed.   At high temperatures re-
action times vary greatly with temperature.   It was
thought that the temperature could be kept within a
total range of 5 - 6°C but it appears that this is not
close enough to obtain consistent results in this type
of work.   Some attempts were made to find if the aus-
tenising time and, in consequence, the degree of homo-
geneity had any effect on the rate of reaction.   Aus-
tenising times ranging from three minutes to one hour
were tried.   The rates of reaction varied but the
variation was no greater than with constant austenis-
ing times and was quite random.   It is impossible to
say from these results that degree of homogeneity does
not produce any effect on reaction rate but it appears
that the effect, if any, is not great.   Roberts and
Mehl reported results showing that lack of homogeneity
in a 0.78% carbon steel could accelerate reaction by
nearly ten times but even from the poor results ob-
tained in these tests it seems very unlikely that any
effect as great as this exists in the case of medium
carbon steels.   The varying austenising times had,

of /

of course, the effect of producing differing austenitic grain sizes but the effect of this did not seem to be as great as would have been expected had the pearlite nuclei been produced at a rate proportional to the austenite grain surface.   It seems probable that the presence of ferrite breaking up the austenite grain alters the relationship entirely.   The rate of growth of the austenite grain size was determined by examination of a series of gradient quenched specimens. The specimens, four inches long, were heated at 900°C for 2, 5, 15 and 30 minutes and were then quenched in water at one end in order to give a range of cooling rates.   The grain size could be determined either from the network of troostitic pearlite produced near the rapidly cooled end where a split transformation had taken place or from the ferrite network produced where the cooling rate was such as to allow the separation of only enough ferrite to form a complete network.   The results are shown in Fig. 14 where the A.S.T.M. grain size is plotted against a logarithmic time scale.   This graph represents the principal grain size present but at the longest time the austenite had not completely grown to the larger size and 20% of the section consisted of finer grains of very variable size which would add greatly to the grain boundary area though the extent of the increase would be difficult to calculate.   The rapid growth of a limited number of grains to produce a mixture of coarse and fine grains is the normal behaviour of a fine /

FIG. 14. AUSTENITIC GRAIN GROWTH.
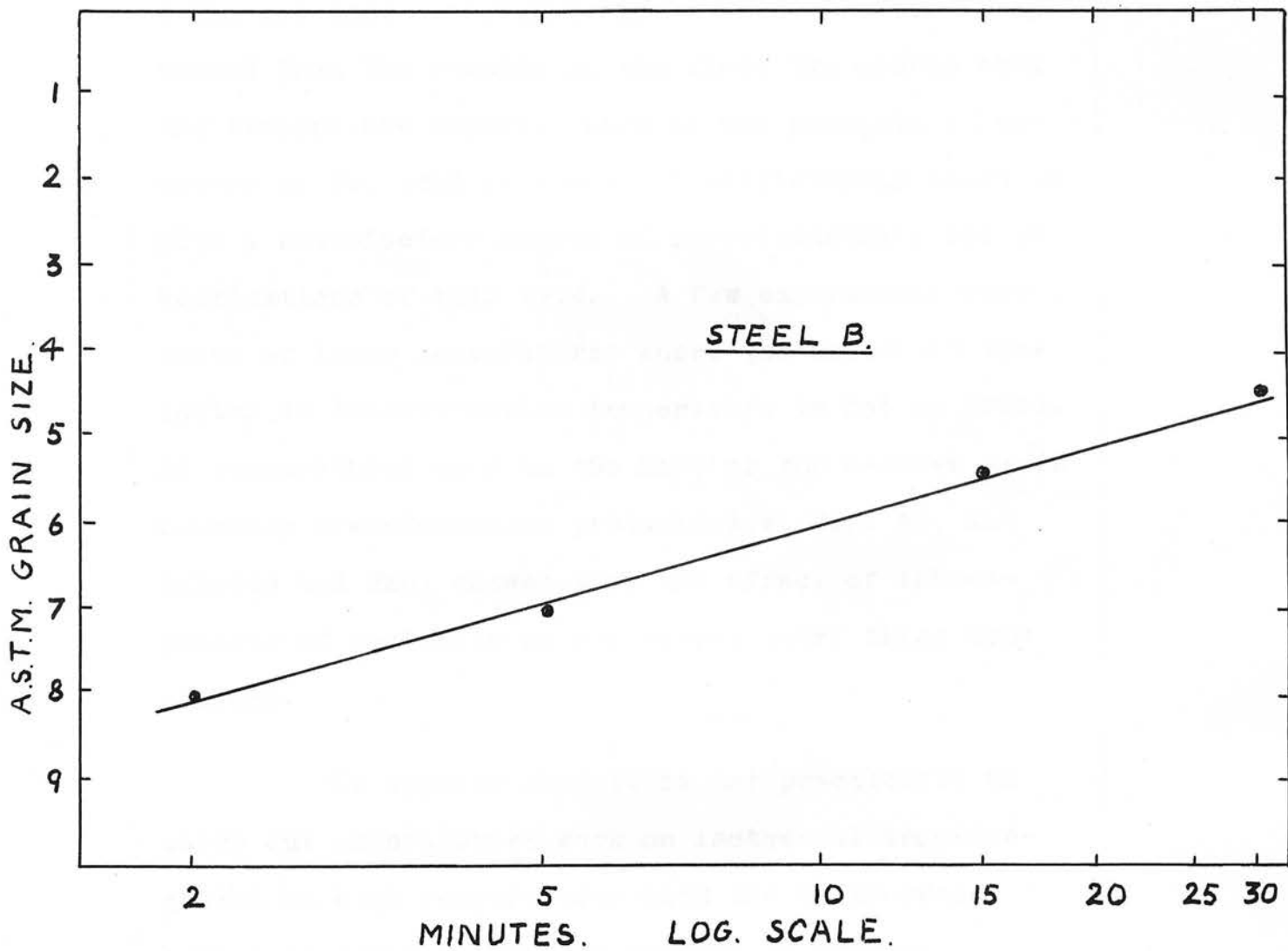
STEEL B.

fine grained steel. On longer heating the smaller
grains are usually absorbed leaving a uniformly coarse
grain but this may require a much longer austenising
time than any industrial heat treatment cycle(4).

Steels A and B were found to behave similarly
as regards isothermal reaction and grain growth. Steel
C was not used as extensively as A and B after it ap-
peared from the results on the first two steels that
the temperature control which it was possible to ex-
ercise on the lead bath was not sufficiently close to
give a satisfactory degree of reproducibility for in-
vestigations of this type. A few experiments were
tried at lower temperatures where the effect of var-
iation in transformation temperature is not so great.
At temperatures near to the knee of the S-curve grain
boundary transformation predominates, Fig. 15, and
Roberts and Mehl showed that the effect of inhomo-
geneity of austenite is not marked under these con-
ditions.

It appears that it is not practicable to
carry out quantitative work on isothermal transfor-
mation at high temperatures with the small scale
equipment available in the average laboratory.
Roberts and Mehl claim that in their experiments the
temperature of the lead bath used was controlled with-
in 0.5°C and it seems that something approaching this
accuracy is necessary for success. It is felt, how-
ever, that although these experiments did not lead to
the /

the results which had been hoped, they did give a con-
siderable amount of qualitative information regarding
the mode of transformation of medium carbon steel.

Normalised Structures.

Micro-examination of normalised specimens
of steel B which had been held for varying times at
the normalising temperature suggested that it might
be possible to determine in this way the time required
for homogeneity to be reached.

In the "as received" condition steel B con-
sisted of a very fine ferrite-pearlite structure,
Fig. 16.   Samples of the steel in this condition
were normalised at 850°C, holding in the furnace for
periods of 5, 10, 20 and 30 minutes before air cooling.
Some of the structures obtained are shown in Figs.
17 - 19.   The most obvious feature is the steady in-
crease in grain size with holding time.   Other speci-
mens were then similarly normalised after having been
previously annealed at 900°C to give the much coarser
structure shown in Fig. 20.   Normalised structures
are shown in Figs. 21 - 23.   In the annealed steel
the carbon would have, on an average, to diffuse much
further than in the "as received" steel in order to
produce homogeneous austenite during normalising and
thus should take longer to become uniformly distri-
buted.   Figs. 17 and 20 show the difference in the
two cases after holding for five minutes.   The "as
received" /
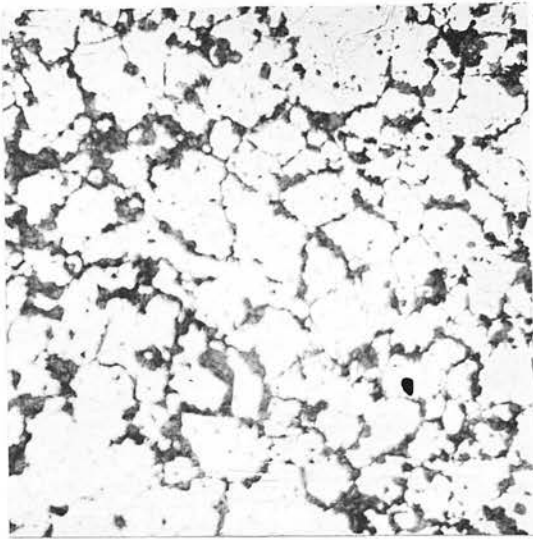
Fig. 15. Steel C. Austenised 800°C
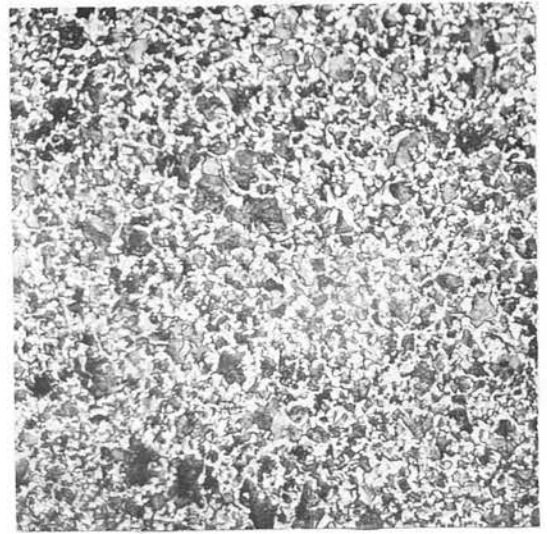Reacted 2 sec. at 600°C.   x 250.

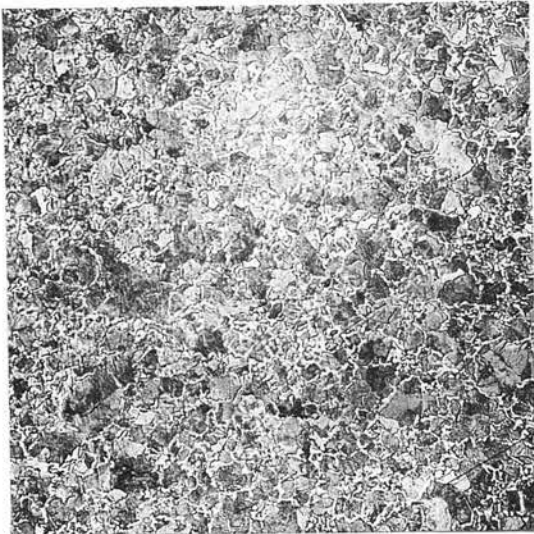Fig. 16. Steel B. as received.
x 250.

Fig. 17. Steel B. Air cooled after
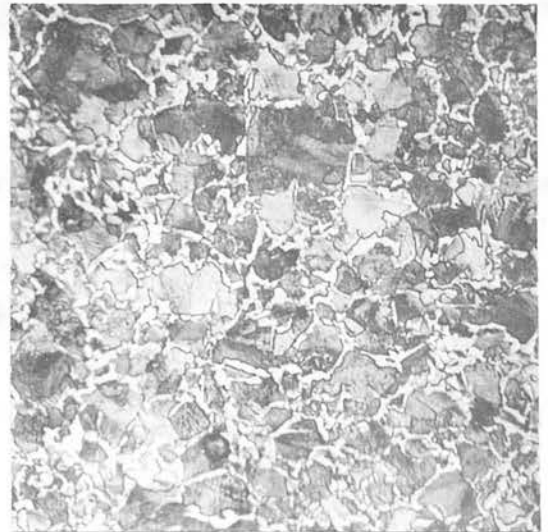5 min. at 850°C   x 250.

Fig. 18. Steel B. Air cooled
after 10 min. at 850°C   x 250.

received" steel has produced a fine, uniform structure while the annealed sample shows a very irregular structure with ragged pearlite areas of varying size. Here and there ferrite patches can be seen in the centre of pearlite areas suggesting that either the large ferrite areas present in the annealed steel were not completely absorbed or that low carbon areas in the austenite resulting from incomplete diffusion had precipitated ferrite at these points during cooling. After ten minutes in the normalising furnace these differences have largely disappeared, Figs. 18 and 22, while after 30 minutes the effect of the preliminary treatment can no longer be seen. These results suggest that inhomogeneity caused by the presence of ferrite in the annealed steel disappears within ten minutes of the insertion of the specimen in the normalising furnace. Since these specimens take about 2 minutes to reach the upper critical temperature in the furnace used, diffusion of carbon is completed, for practical purposes in not more than eight minutes of the piece reaching the normalising temperature and possibly in appreciably less. Experiments described later suggest that the time is less.

A series of tensile tests was carried out, through the courtesy of Professor W. R. D. Jones, D.Sc., University College of South Wales and Monmouth, on pieces heat treated in the same way as the above micro-specimens in order to find if they showed any effect. /
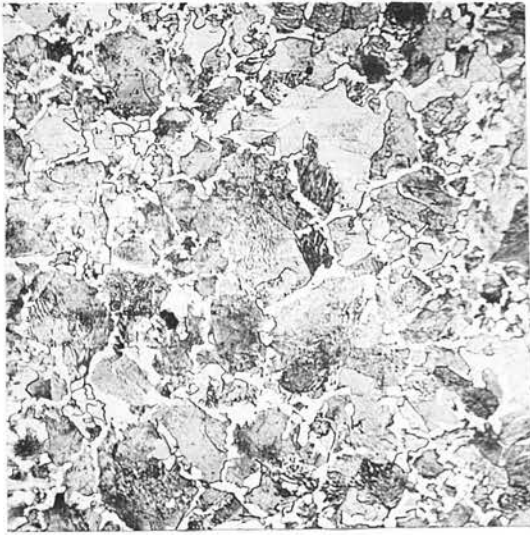
Fig. 19. Steel B. Air cooled
   after 30 min. at 850°C. x 250.

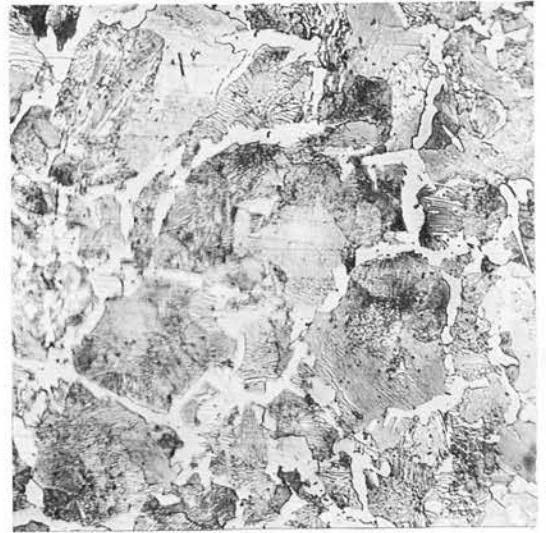Fig. 20. Steel B.
   Annealed   x 250.
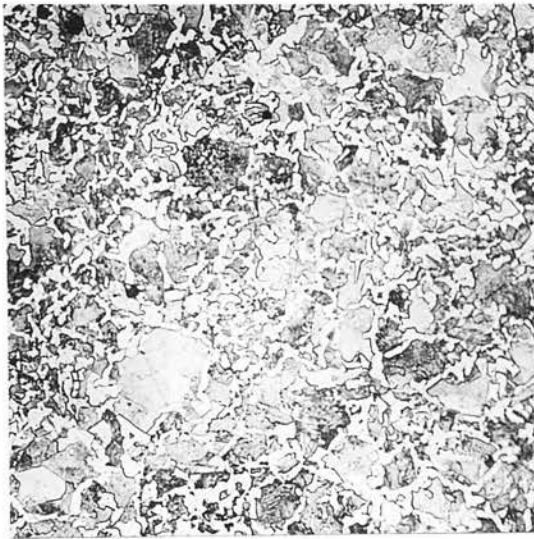

Fig. 21. Steel B. Air cooled
   after 5 min. at 850°C.   Pre-
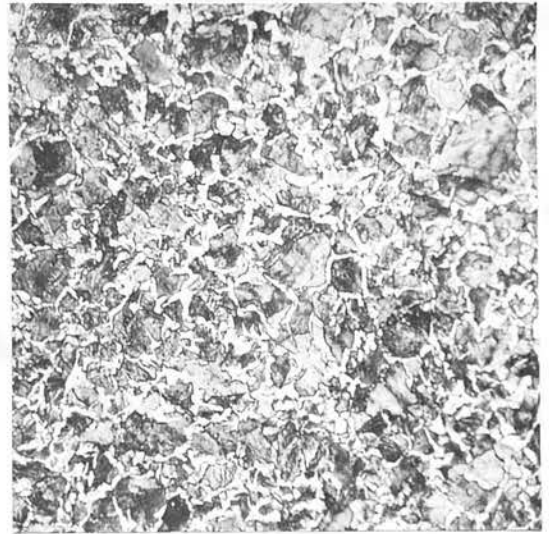   viously annealed.   x 250.

Fig. 22. Steel B. Air cooled
   after 10 min. at 850°C. Pre-
   viously annealed.   x 250.

effect parallel to that observed visually.   Two
Tensometer tensile test pieces were machined from
each of the heat treated bars.   The cross section of
the test pieces used was 0.02 sq. in. and the gauge
length 0.6 in.   The results on the individual test
pieces and the average results are shown in Table III.
Both "as received" and annealed specimens show a pro-
gressive increase in maximum stress and a tendency for
the ductility to decrease with increasing normalising
times.   Differences between "as received" and anneal-
ed specimens for the same normalising time are probably
not significant.   The rod used is too thin to allow
impact tests or Tensometer slow bend tests to be car-
ried out on hardened specimens.

<div align="center">TABLE   III   /</div>

TABLE III.

Tensometer Tensile Tests on Steel B.

Initial State	Normalising Time Mins.	Yield Point Tons/ sq.in.	Maximum Stress Tons/ sq.in.	Elongation %	Reduction of Area %
As received	5	31.1) 30.5) 30.8	46.7) 45.0) 45.9	24) 27) 26	45) 43) 44
	10	29.4) 30.1) 29.8	47.8) 47.9) 47.9	25) 23) 24	44) 43) 44
	20	29.5) 30.6) 30.1	48.4) 48.5) 48.5	22) 23) 23	43) 43) 43
	30	28.4) 29.2) 28.8	48.7) 49.4) 49.1	24) 24) 24	43) 45) 44
Annealed	5	29.5) 29.2) 29.4	47.0) 46.1) 46.6	28) 27) 28	48) 48) 48
	10	29.2) 29.7) 29.5	48.1) 47.1) 47.6	24) 23) 24	44) 44) 44
	20	29.2) 28.8) 29.0	48.4) 47.8) 48.1	22) 26) 24	43) 44) 44
	30	29.2) 28.7) 29.0	48.6) 48.4) 48.5	26) 22) 24	45) 43) 44

End Quench Tests.

Further information regarding the manner in which the breakdown of austenite is affected by inhomogeneity was obtained from end quench tests.

The form in which the steels being used were available ruled out the use of the standard Jominy end quench test but a simple form of test was used which gave the required continuous variation in cooling rate. The /

The apparatus consisted only of a large glass funnel
discharging through a length of rubber tube and a jet
of $\frac{1}{4}$" internal diameter glass tube about 18" below the
level of the funnel.   The height of the water jet was
adjusted to 1" by a screw clip on the rubber tube and a
spring clip allowed the water to be turned on without
delay.   A jig made of thin asbestos sheet allowed the
specimen, supported by the wire on which it hung in the
furnace, to be placed accurately above the jet with its
lower end half an inch above the outlet.   The specimens
used were $3\frac{1}{2}$" long.

After quenching, about $1\frac{1}{2}$ in. at the quenched
end of the test piece was ground on the side of a wheel
so as to produce a flat about 1/8 in. wide.   This
piece was then cut off and prepared for micro-examination.
The use of a grinding wheel, even with careful cooling
of the specimen, tempered the martensite present to some
extent and the effect was clearly visible under the micro-
scope.   Hand grinding on a flat carborundum stone gave
excellent results with no visible signs of tempering
but was much slower and more laborious and was, as a
rule, unnecessary.   Hand grinding was, however, used
in all cases, not only of end quench tests, in which
much metal had to be removed prior to a careful examin-
ation of a steel containing much martensite.

Steels A and B when tested in this way showed
a fully martensitic structure for just over half an
inch from the end of the test piece.   Metal more dis-
tant /

distant from the end showed a partial transformation to troostitic pearlite and as the distance increased so did the proportion of pearlite. Within about another eighth of an inch after the first appearance of pearlite the structure was wholly pearlitic and at greater distances increasing amounts of ferrite appeared. The same general pattern appeared irrespective of austenising time or temperature. The cooling rate evidently changed very sharply with distance from the end of the test piece and changes in critical cooling rate due to changes in grain size or other factors did not show up as measurable changes in the distance from the end of the test piece at which definite structures were formed. With very short austenising times, however, certain features appeared in the structures which seemed significant. Figs. 24 and 25 show the normal appearance of the zone of split transformation, produced in this case in steel B after austenising at 850°C for 5 minutes. Transformation begins at isolated points. Further from the end of the test piece, at a somewhat lower rate of cooling, transformation takes place round the austenite grain boundary forming a network of troostitic pearlite which can be used to determine the austenitic grain size. With very short austenising times the ferrite originally present in the steel is not completely absorbed by the austenite and the residual ferrite can be seen in the martensite produced when the steel is quenched. Fig. 26 shows residual ferrite in the martensite zone of an end quench test piece after austenising for 2 minutes, only /
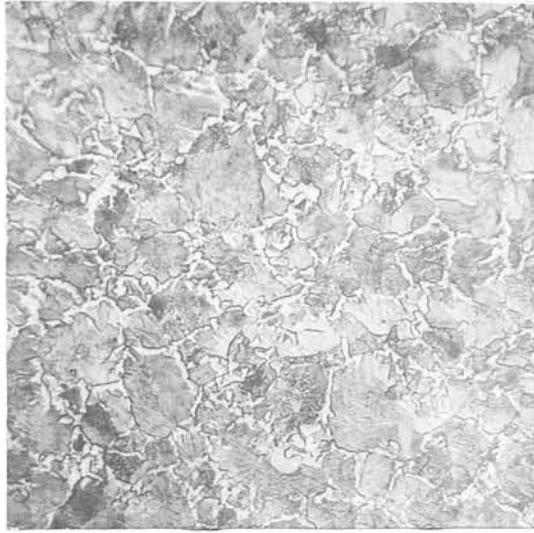
Fig. 23. Steel B. Air cooled
after 30 min. at 850°C. Pre-
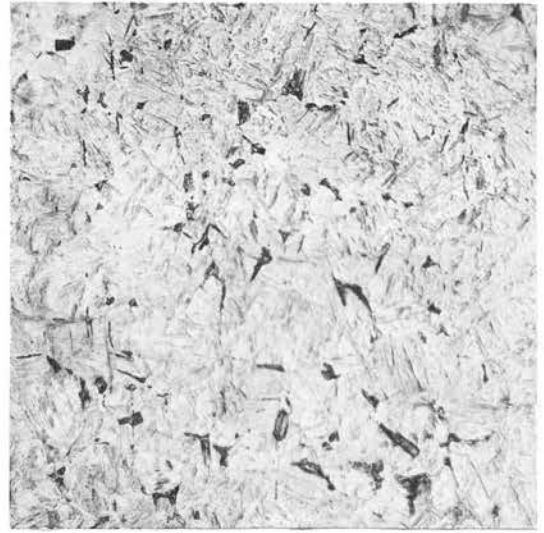viously annealed. x 250.
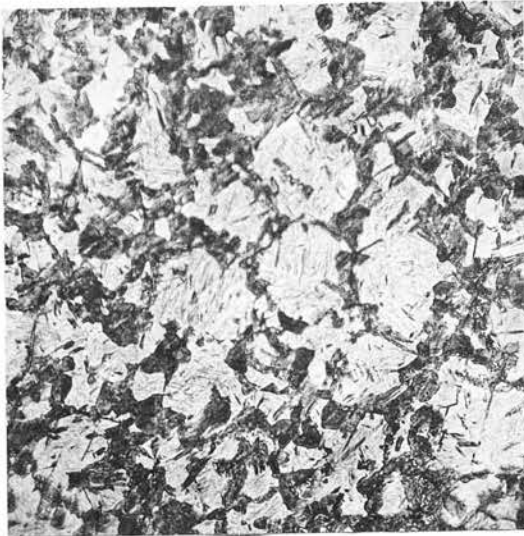

Fig. 24. Steel B. End
quenched. x 500.

Fig. 25. Steel B. End
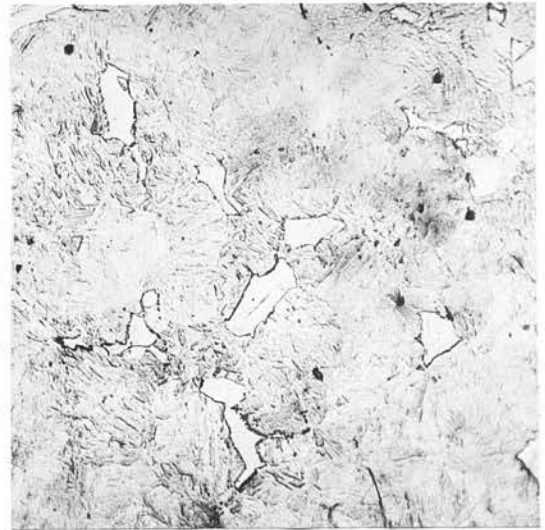quenched. x 500.

Fig. 26. Steel B. Austenised
2 min. End quenched. x 500.

only just long enough to allow the test piece to be
brought up to, or a little above, the upper critical
temperature.   This ferrite remained from the initial
coarse annealed structure shown in Fig. 27.   When trans-
formation began in this case it did not begin at isolated
points distributed apparently at random but began round
the edges of the residual ferrite and at no other points.
There are two possible explanations of this.   The first,
that pearlite is nucleated by ferrite, can at once be
ruled out.   There is abundant evidence that pearlite
is nucleated by cementite and not by ferrite (5).   The
other explanation, the existence round the residual
ferrite of a region in which the carbon content is lower
than in the rest of the austenite and which has, there-
fore, a higher critical cooling rate, seems to be the
only possible one.   At slightly greater distances from
the end of the test piece where the cooling rate is less,
the pearlite formation extends beyond the ends of the
ferrite areas and begins to form a coarse network, Figs.
28 and 29.   This network corresponds in size with the
ferrite network of the original coarse annealed struc-
ture shown in Fig. 27.   It appears that though the
ferrite has been completely absorbed over most of the
original ferrite net work the carbon has not diffused
sufficiently to even out the carbon content of the aus-
tenite and the position of the original ferrite is still
marked by regions of somewhat lower carbon content.   The
material in these regions has, thus, a higher critical
cooling rate and the difference is sufficiently marked
to /

Fig. 27.   Steel B.   Annealed.
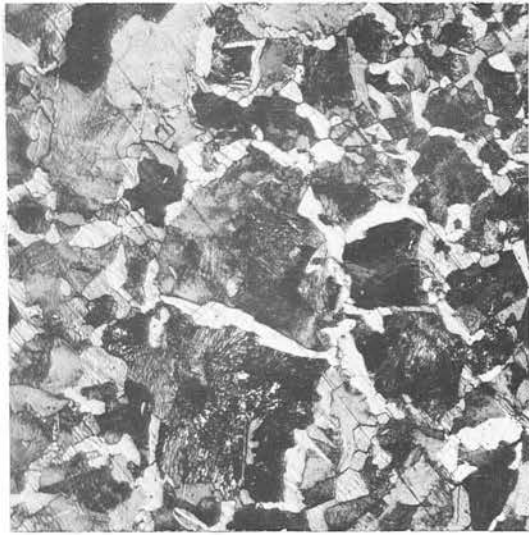   x 500.

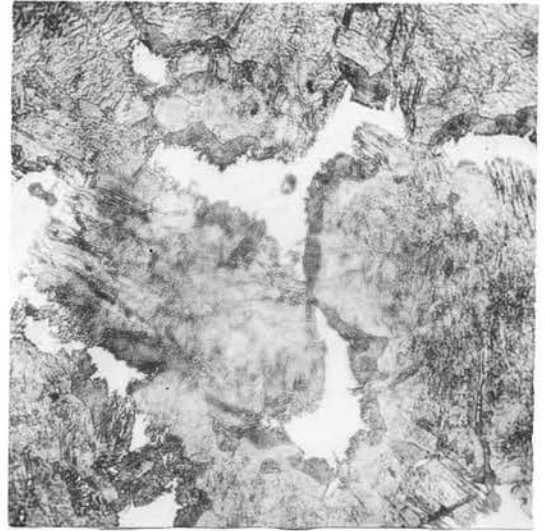Fig. 28.   Steel B.   Austenised
2 min.   End quenched   x ~~500.~~
                                          1,000
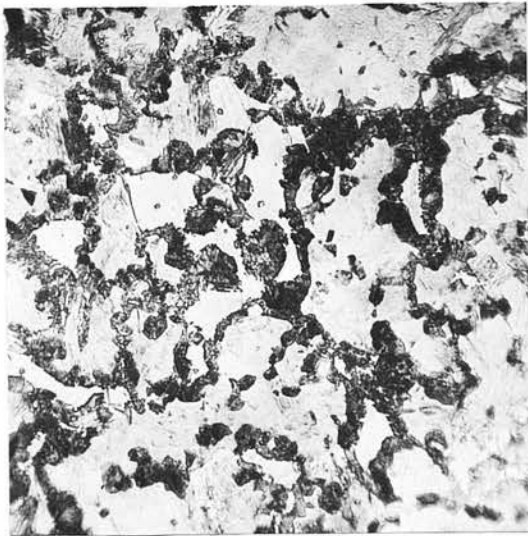

Fig. 29.   Steel B.   Austenised
2 min.   End quenched   x ~~4,000.~~
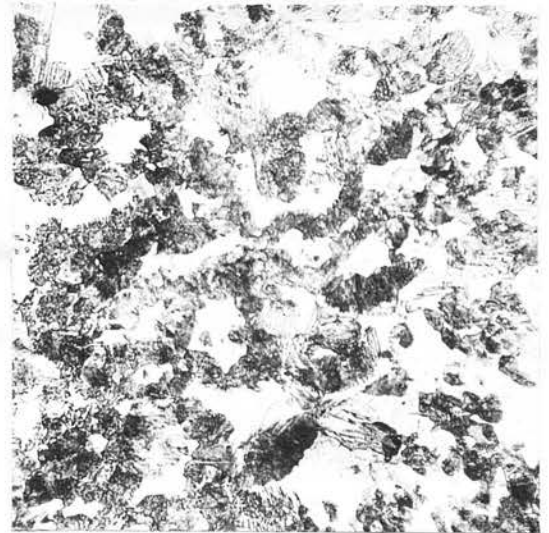   500

Fig. 30.   Steel B.   Austenised
2 min.   End quenched.   x 500.

these experiments than has any other factor and that
this inhomogeneity is eliminated in the first few min-
utes after the specimen reaches the heat treatment tem-
perature.   The time required for heat treatment of a
piece of steel appreciably larger than the small test
pieces employed in these experiments is determined by
the rate of heating and not by the time required for
structural changes to take place.   The possibility is
not ruled out that undissolved carbide may be present
even after the inhomogeneity due to ferrite is elimin-
ated and that the gradual solution of this carbide de-
lays the start of the transformation without altering
the appearance of the structures produced but causing
them to be produced further from the quenched end of
the test piece.   This seems unlikely since cementite
is a powerful innoculant for pearlite and yet, after
short austenising times when cementite particles would
be most abundant, carbon concentration gradients are the
determining factor.   The specimens before treatment con-
tained the carbide in the form offairly fine pearlite
and it appears that this is completely absorbed before
the effect of the ferrite is entirely removed.   If the
initial structure of the steel was such that cementite
was present in larger particles then the picture would
be entirely different and it would probably be found
that the time required for carbide solution was the
determining factor.   The presence of a coarser ferrite
structure would increase the time required for diffusion
at any given temperature but this time would still be
the /

the factor determining the time of soaking required.

Conclusion and Summary.

The work described in this section is a study of one aspect of the effect of austenitic inhomogeneity on the transformation of austenite during the heat treatment of plain carbon steels containing about 0.4% carbon. To begin with it was intended to make a study of the austempering process and it was shown that the change in electrical resistance accompanying the precipitation of carbide could be used to follow the breakdown of austenite. The use of this method meant that a non-conducting quenching bath must be used since it was not feasible to provide an insulating cover for the test piece. Salt bath and hot air blast quenching were both tried but proved to be too slow to supress completely the formation of pearlite. Attention was then transferred to a more direct study of the effect of inhomogeneity by microscopical methods. Specimens were reacted isothermally in an attempt to detect any differences in reaction times due to differing degree of homogeneity. The results were inconsistent, due probably to variations in the temperature of the lead bath used, and no correlation between austenising time and reaction time could be established. Differences in the appearance of specimens air cooled after varying times in the furnace suggested that a comparison of the appearance of normalised specimens which had initial structures of varying /

varying degrees of coarseness might give an approximate idea of the time required to eliminate the effect of varying initial structure. The specimens employed had the carbide in the form of pearlite of approximately the same degree of fineness but the ferrite areas differed in size, one having an ASTM grain size 2-3 and the other 8. It was found that after three minutes at 850°C followed by air cooling the structures obtained showed marked differences but that after eight minutes no difference was visible. It seems, therefore, that between three and eight minutes carbon has diffused sufficiently to remove the effect of the ferrite corresponding to the coarser grain. No conclusion can be drawn regarding the time taken for solution of carbide. Tensile tests on specimens prepared in the same way showed differences after three minutes at temperatures which were markedly reduced after eight minutes, supporting the conclusion above. End quenched test pieces showed the effect of inhomogeneity due to ferrite on transformation during more rapid cooling. In this case the presence of areas of low carbon content was shown to result in the formation of pearlite at cooling rates too great to allow of its formation by the normal grain boundary nucleation. Starting with a ferrite-pearlite structure with grain size 4, the low carbon areas resulting from the ferrite in the original structure are eliminated after two minutes at 850°C, at least to the extent that they no longer act more powerfully than the normal grain boundary nuclei in inducing the formation of pearlite.

No /

No conclusion can be drawn with any degree of condifence
regarding the time required to dissolve cementite com-
pletely and remove the resulting inhomogeneity but it
seems probable that it is less than the time required
to remove inhomogeneity due to the original presence
of ferrite.   This would not be the case if the original
structure contained cementite in larger particles than
was the case in these experiments.

The practical conclusion to be drawn from
these experiments in that in the heat treatment of 0.4%
to 0.5% carbon steel where the carbon is present in the
form of sorbitic or finely lamellar pearlite, the time
required at a given temperature for practical homogene-
ity to be reached is determined by the time required
for carbon to diffuse into the pre-existing ferrite
areas.   This time depends on the size of those areas
but is quite short, of the order of two minutes at
850°C where the ferrite network corresponds to A.S.T.M.
grain size 4.   Unless the piece being heat treated is
very small, the time required for homogenisation will
be so short compared with the time required to raise
the temperature that in most cases it will be negligible.

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