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An Isothermal Study of an S.A.E. 6150 Steel

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AN ISOTHERMAL STUDY OF AN
S.A.E. 6150 Steel

by

Tom Liss

in conjunction with

C. Arentzen

A Thesis

submitted to the Department of Metallurgy

in Partial Fulfillment of

the Requirements for the Degree of

Bachelor of Science in Metallurgical Engineering

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INTRODUCTION

PURPOSE:

The purpose of this thesis is to set forth the method followed, the laboratory procedure practiced, the results obtained, the conclusions drawn, and the recommendations proposed as a consequence of a metallographic study of the isothermal transformation of a S.A.E. 6150 steel.

The results that were obtained from a magnetic permeability method of isothermal study of the same steel are compared with those that were obtained in the course of this work.

HISTORY OF PROCESS:

In 1929 two brilliant metallurgists, E.C.Bain and E.S.Davenport,(1) presented to the A.I.M.E. a paper that dealt with isothermal transformation studies of numerous steels. Prior to this date, the investigation of the properties that can be obtained from the various heat treating procedures to which a steel can be subjected dealt primarily with the composition of the material, or the cooling velocity of the quench. The early investigators, by varying the composition of the steels under test, made allowance for one variable of the heat treating process, but since they employed continuous quenching techniques, they failed to differentiate between the separate effects that are due to time or to temperature.

Portevain and Garven(2) investigated a transformation that takes place as a consequence of continuous cooling. The theory that governs this transformation will be discussed later in the thesis. But the true process of isothermal transformation remained hidden until Bain and Davenport made their studies.

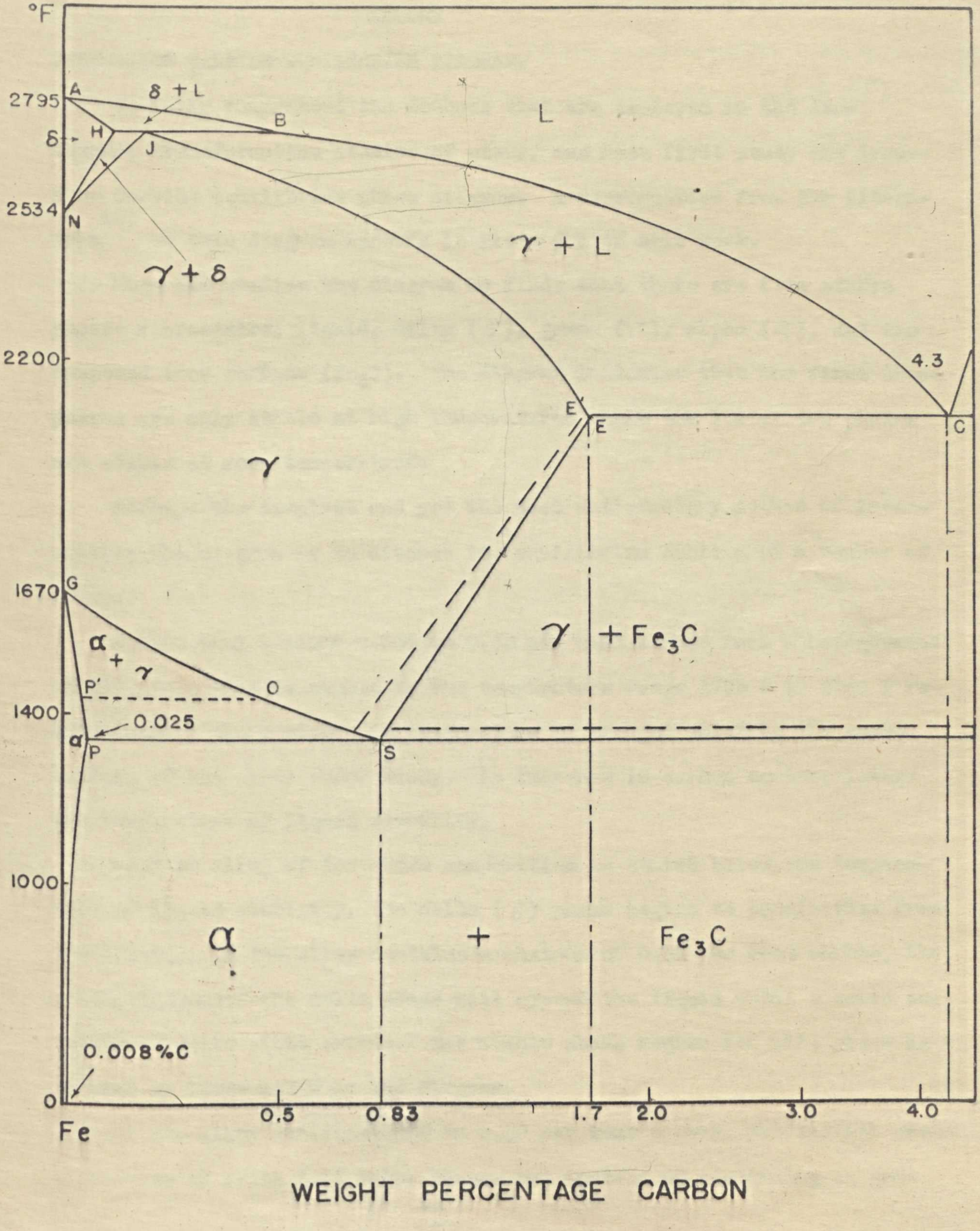
After the 1929 meeting of the A. I. M. E., the work of Bain and

Note: The superscripts in brackets refer to the bibliographic list given on page 43.

Davenport was substantiated by various investigators, and as a consequence, additional information was added to the process of heat treating. The investigations are still being conducted and at present the results of isothermal studies that have been conducted on the numerous steels that are used in the heat-treating processes are fully appreciated by the industry.

PLATE No. 1

IRON-CARBIDE DIAGRAM



THEORY

IRON--IRON CARBIDE EQUILIBRIUM DIAGRAM:

To fully comprehend the methods that are employed in the isothermal transformation studies of steel, one must first study the Iron--Iron Carbide equilibrium phase diagram. A reproduction from the literature (3) of this diagram appears in Plate # 1 of this work.

When one studies the diagram he finds that there are five stable phases represented, liquid, delta (δ), gamma (γ), alpha (α), and the compound iron carbide (Fe_3C). The diagram indicates that the first three phases are only stable at high temperatures while the latter two phases are stable at room temperature.

Perhaps the simplest and yet the most satisfactory method of interpreting the diagram is to discuss the equilibrium cooling of a number of alloys.

Alloys that contain 0.000 to 0.55 per cent carbon form a homogeneous liquid phase that is stable in the temperature range 2795 F to 2720 F respectively. The variance in temperature is brought about by the carbon content of the alloy under study. An increase in carbon content lowers the temperature of liquid stability.

When an alloy of the above composition is cooled below the temperature of liquid stability, the delta (δ) phase begins to precipitate from the liquid. If the alloy contains a maximum of 0.08 per cent carbon, the precipitation of the delta phase will expend the liquid until a solid solution of delta alone exists. The stable phase region for this phase is bounded by lines A-H-N in the diagram.

If the alloy contains 0.08 to 0.55 per cent carbon, the initial precipitation of delta (δ) takes place, but instead of continuing to grow at the expense of a melt until a simple solid solution phase exists a peritectic reaction occurs at 2710 F. If the alloy contains 0.08 to 0.18 per cent carbon, the product of this reaction will be gamma (γ) iron plus the primary delta iron. Alloys in the range 0.18 to 0.55 per cent iron will react to give gamma iron plus liquid, and finally if the alloy contains exactly 0.18 per cent carbon the product of the reaction will be gamma iron.

Below a temperature of approximately 2535 F any alloy in the 0.000 to 0.55 per cent carbon range will be in the gamma condition. This phase is a solid solution of carbon in face centered cubic iron.

The gamma phase in the case of alloys that contain a maximum of 0.55 per cent carbon is stable in the temperature range 1670 F to 1400 F respectively. Again we find that the higher the carbon content of the alloy the lower the temperature of stability in the gamma phase region.

Below the critical temperature a new structure, alpha (α) iron, will precipitate from the solid gamma solution. Alpha iron is a solid solution of carbon in iron that has a body centered cubic structure.

The precipitation of alpha (α) from gamma (γ) continues until the temperature 1333 F is attained. At this temperature a eutectoid reaction takes place, and the remaining gamma iron, that consists of an alloy of 0.80 carbon, simultaneously precipitates alpha iron and the intermetallic compound, iron carbide (Fe_3C). The eutectoid structure is stable to room temperature.

A slight total solubility of carbon in alpha iron is shown by the phase region bounded by lines Q-P-G. The maximum solubility of this phase occurs at the eutectoid temperature and it gradually decreases until a solubility of 0.008 per cent is found at room temperature.

The next alloy division is those alloys that contain 0.55 to 0.83 per cent carbon. These alloys form a stable liquid phase in the temperature range 2710 to 2680 F. Below this temperature, gamma iron will precipitate from the liquid. These alloys do not experience the peritectic reaction; instead, they form the solid solution of gamma iron on the solidification of the liquid. The reactions that the gamma solid solution experiences on cooling to room temperature are identical to those described for the previous alloys.

Alloys in the range 0.83 to 1.7 per cent carbon have the same initial solidification reactions as those of 0.55 to 0.83 per cent carbon, but when they experience the eutectoid reaction at 1333 F, primary iron carbide (Fe_3C) is precipitated from the gamma (γ) solid solution instead of primary alpha iron. The eutectoid reaction is identical to the previously mentioned alloys. The structure is again stable down to room temperature.

Alloys in the range 1.7 to 6.67 per cent carbon (the 6.67 per cent alloy is the extent of the usefully commercial alloys; therefore, it terminates the diagram.)follow a simple eutectic type diagram on primary solidification. The eutectic point occurs at 4.3 per cent carbon and at 2075 F. The primary phase that precipitates from the melt will either be gamma iron or iron carbide. The composition of the primary phase will be determined by the composition of the alloy--1.7 to 4.3 per carbon alloys give primary gamma; 4.3 to 6.67 yield iron carbide (Fe_3C); the 4.3 alloy would be 100 per cent eutectic.

The gamma that results either from a primary precipitation or from the eutectic reaction is stable down to 1333 F where the previously discussed eutectoid reaction occurs. After this reaction is completed, the alloy is stable to room temperature if the decrease in solubility of carbon in (δ) iron is disregarded.

In the course of the many years that have been devoted to the study of the equilibrium diagram, a more common method of naming the phases present in the given phase regions has been adopted. Gamma iron is called austenite. Ferrite is the common nomenclature applied to alpha iron. The intermetallic compound (Fe_3C) is named cementite, and the eutectoid of iron carbide and alpha iron is termed pearlite.

In extending nomenclature to the diagram, alloys containing up to 1.7 per cent carbon are termed steel, and those containing 1.7 to 6.67 per cent carbon are called cast iron. There has been a growing tendency to classify the alloys of 1.7 to 4.3 per cent carbon as semi-steels, but this classification is not common.

The true steels are subdivided into two divisions. Those below eutectoid composition are termed hypo-eutectoid steel; hyper-eutectoid is the designation applied to the steels of composition 0.83 to 1.7 per cent carbon.

In the preceding discussion equilibrium cooling yielded a final product of either primary alpha iron or iron carbide surrounded by a eutectoid of alpha iron plus iron carbide. But if non-equilibrium techniques are used to cool the alloys from the gamma region to room temperature, the final product that will exist at room temperature may exhibit a very different microstructure.

Grange and Kiefer (4) have studied the microstructures that result from non-equilibrium cooling. They found that cooling rates of less than 40 F per hour will produce microstructures that duplicate equilibrium cooling microstructures; cooling rates of 40 to 150 F per hour will produce new constituents in the microstructure. The new structures are acicular in appearance, and share the microstructure with products that are formed by equilibrium cooling.

Cooling rates of 150 to 2100 F per hour will produce a microstructure that contains three phases; two of the phases are acicular and are difficult to resolve. One can distinguish between them by the different characteristics they exhibit when polished and etched. The third constituent is proeutectoid ferrite.

When the cooling rate is increased in the range of 150 to 2100 F per hour, the proeutectoid ferrite does not separate from the austenite but, the two acicular structures are the only constituents of the microstructure. One of the acicular structures is very fine, highly irresolvable pearlite. The other constituent is an entirely new phase commonly termed martensite. It is a body centered tetragonal structure of a solid solution of carbon in iron. The mechanism of martensite formation from the austenite is fully described on page 263 of the American Society for Metals Handbook for 1948 (3), and the reader is referred to the publication.

Martensite is the structure that contributes the property of high hardness to a steel when it is quenched very rapidly from the gamma region. To form a homogeneous martensite structure one must employ quenches of at least 54,000 F per hour.

THEORY OF ISOTHERMAL TRANSFORMATION:

As was previously mentioned, the early investigators were concerned primarily with the effects of quenching velocity. They knew that definite microstructures could be produced by quenching at certain predetermined rates, but as explained in the previous discussion, many quenching velocities did not yield homogeneous microstructures.

Bain and Davenport reasoned that the heterogeneous microstructures were the result of transformations taking place at varying temperatures during the quenching operation, and that if a given steel could be allowed to transform isothermally a homogeneous microstructure would result.

To prove this reasoning, they conducted isothermal studies for various steels. In these tests they employed the following techniques:

1. A large number of small samples were prepared.
2. The samples were heated in a furnace whose temperature was maintained above the upper critical point as shown in the equilibrium diagram.
3. The samples were quenched into a bath whose temperature was below the critical temperature. This bath provided an isothermal system for the transformation.
4. At varying time intervals a sample was removed from the bath and rapidly quenched into a water bath.
5. Hardness measurements were made on the various samples, and the microstructure was investigated by metallographic methods.

Since rapid quenching, such as a water bath would provide, transforms austenite to martensite, a measure of the austenite that did not transform isothermally could be measured in terms of martensite observed in the sample.

An isothermal graph of per cent martensite versus time of transformation is shown in the figure in plate # 2.

When Bain and Davenport evaluated the results of an isothermal study they found that their reasoning had been substantiated. Transformation did take place isothermally, and when complete, a homogeneous microstructure resulted.

The method that one uses to combine a large number of isothermal graphs into one chart is perhaps best understood when one studies plate #2;

PLATE No. 2

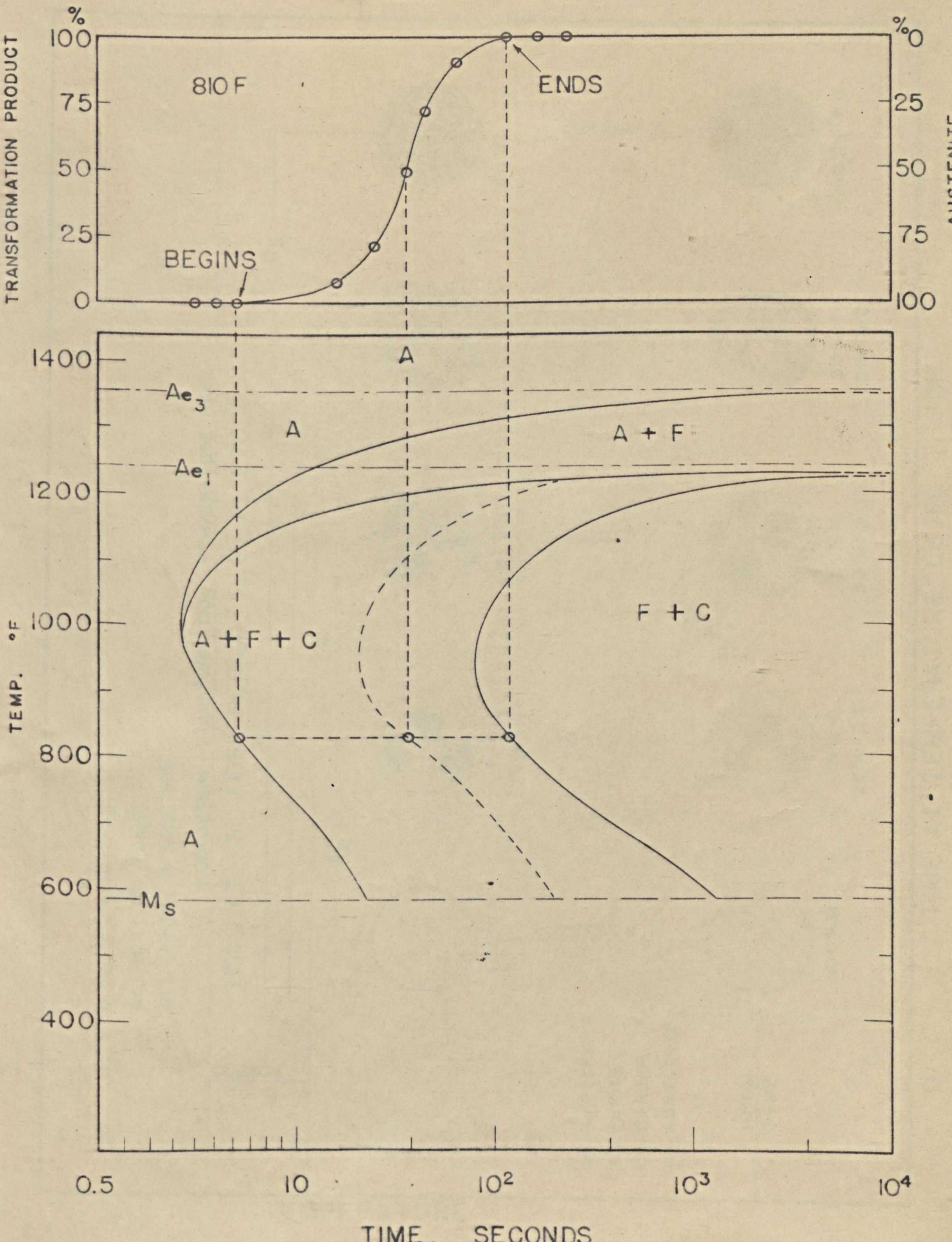
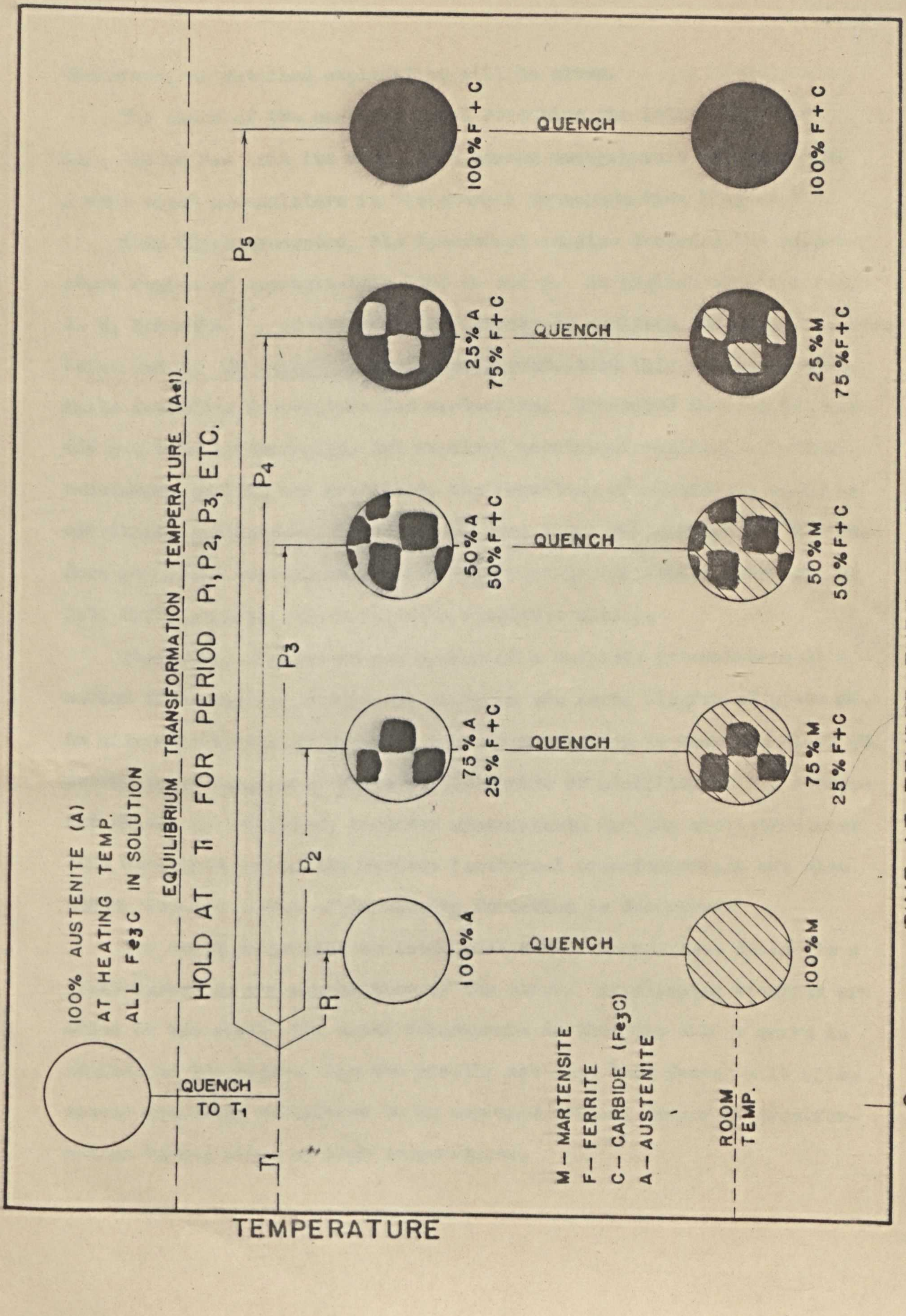

%
TRANSFORMATION PRODUCT
100
75
50
25
0
%0
25
50
75
100
AUSTENITE
810 F
BEGINS
ENDS
1400
1200
1000
800
600
400
TEMP. °F
Ae3
Ae1
Ms
A
A
A + F
A + F + C
F + C
A
0.5
10
10²
10³
10⁴
TIME, SECONDS

PLATE No.3


100% AUSTENITE (A)
AT HEATING TEMP.
ALL Fe3C IN SOLUTION
QUENCH
TO T1
EQUILIBRIUM TRANSFORMATION TEMPERATURE (Ae1)
HOLD AT T1 FOR PERIOD P1, P2, P3, ETC.
T1
P1
P2
P3
P4
P5
100% A
75% A
25% F+C
50% A
50% F+C
25% A
75% F+C
100% F + C
QUENCH
QUENCH
QUENCH
QUENCH
QUENCH
100% M
75% M
25% F+C
50% M
50% F+C
25% M
75% F+C
100% F + C
M — MARTENSITE
F — FERRITE
C — CARBIDE (Fe3C)
A — AUSTENITE
ROOM TEMP.
O
TEMPERATURE
TIME AT TEMPERATURE LEVEL T1 FOR TRANSFORMATION

therefore, no detailed explanation will be given.

The shape of the combined graph resembles the letter "S", and this letter has lent its name for a common nomenclature of the graph. A more exact nomenclature is "Isothermal Transformation Diagram."

When first presented, the isothermal studies included the temperature region of approximately 1400 to 200 F. An English metallurgist, J. G. Robensen (1), questioned the validity of applying isothermal studies below 550 F. He believed, and later proved, that this was the approximate formation temperature for martensite. He stated that martensite did not form isothermally, but required continuous cooling, and when continuous cooling was practiced, the formation of martensite could be considered spontaneous. These views were later substantiated. The before mentioned article in the American Society for Metals Handbook for 1948 fully explains the martensite formation theory.

Therefore, the modern conception of a complete presentation of a number of isothermal studies is shown on the lower diagram of plate #2. An alternate method of graphing the transformation is shown on plate #3. In the lower diagram of plate #2 the fields of stability of the different phases is indicated; hardness measurements for the microstructures that were produced by the various isothermal transformations are also shown, and the region of martensite formation is described.

The exact shape that an isothermal curve assumes does depend to a great extent on the composition of the steel. As alloying elements are added to the steel, the usual consequence is that the entire curve is shifted to the right. One can readily see that this factor will allow slower quenching velocities to be employed without danger of transformation taking place at high temperatures.

Another factor that will slightly change the shape of the curve is the grain size of the gamma grains. The larger the grains, the slower the transformation for any given temperature.

To conduct an isothermal study, one can apply any of the conventional methods that would differentiate a change in crystal structure. A few would be:

1. Electrical Conductivity
2. Metallographic Investigation
3. Dilatometric Investigation
4. Magnetic Permeability (Applies to ferrous alloys)

SIGNIFICANCE OF ISOTHERMAL DIAGRAMS

One should fully realize that although isothermal diagrams offer a key with which one can open the door to the answers to many of the perplexing problems of heat treating, they also have vefy definite limitations. A very complete discussion of the uses and limitations can be found in a handbook that is published by the Republic Alloy Steel Company. (5) The following summary is taken from this work:

USES:

1. By correolating isothermal diagrams and continuous cooling diagrams, one can arrive at a first approximation of the type of microstructures that will be formed at given cooling rates, and the hardness of the microstructures.
2. Annealing cycles can be calculated from isothermal diagrams.
3. The temperature region of martensite formation can be established.
4. If the size of the sections undergoing heat treatment is such that isothermal cooling will apply to all parts of the section, specialized methods of annealing, and tempering can be established. These methods are fully enumerated in Republic Alloy booklet on pages 167 to 169.
5. They yield valuable information on the theory of nucleation and diffusion.

LIMITATIONS OF DIAGRAM:

1. Most diagrams that are found in the literature represent the average values. They can not be applied directly to any given steel.
2. Small variations in grain size or austenitizing temperature will vary the diagram.
3. The usual shapes and sections that are subjected to commercial heat treating practices are larger than the specimens that were used in the preparation of the isothermal diagram; therefore, they do not respond to isothermal cooling throughout the section. Consequently, isothermal techniques can not be applied to them.

LABORATORY PROCEDURE

PRELIMINARY INVESTIGATION:

The material for this investigation was an S.A.E. 6150 steel that was obtained from the Great Lakes Steel Company of Detroit, Michigan. The results that were obtained from a metallographic investigation of the material indicated that its microstructure was that of a normalized structure. Plate #7 is a reproduction of the structure.

Since steel is usually in the annealed condition before it is heat treated, the samples were box annealed. A reproduction of the annealed structure appears in plate #8.

CHEMICAL ANALYSIS:

The chemical composition of the steel as furnished by the Great Lakes Steel Company was:

Carbon.................	0.48%
Chromium...............	0.93%
Manganese..............	0.77%
Vanadium...............	0.14%

The limitations in variance of composition as given by the Society of Automotive Engineers and American Iron and Steel Institute on page 308 of the American Society for Metal Handbook for 1948 are:

Carbon.................	0.45%----0.53%
Chromium...............	0.80%----1.10%
Manganese..............	0.70%----0.90%
Vanadium...............	0.15% Minimum
Sulfur.................	0.040% Maximum
Phosphorous............	0.040% Maximum
Silicon.....	0.20%----0.35%

GRAIN SIZE:

The grain size determination was made by the McQuaid-Ehn method as described in Kehl's "Metallographic Laboratory Handbook",(6) The counting of the grains was conducted with the aid of a grain size counting eyepiece. The results of this determination established the grain size of this steel as between grain size #7 and #8 as given by the A.S.T.M. E 19 chart. This would be approximately 96 grains per square inch.

ISOTHERMAL TESTING:

The equipment that was used in the isothermal studies is shown in plate #4. It consists of two furnaces with their auxiliary resistence controls. A water bath was used for the quenching medium.

The size of the samples used for the isothermal studies was approximately 1/2 in. by 1/2 in. by 1/8 in. A small hole was drilled in the samples and a wire inserted into the hole. The complete device for handling the samples is shown in plate #5.

Prior to isothermal testing, the time and temperature that was needed to produce a homogeneous gamma phase was established by the following methods:

1. Five of the small samples were placed in the austenitizing furnace. The temperature of the furnace was 1650 F.

2. At intervals of 3, 5, 10, 15, and 20 minutes one sample was removed from the furnace and quenched in cold water.

3. Hardness measurements, and metallographic investigations were utilized to determine the time necessary to produce a completely martensitic structure.

The results of the test showed that completely martensitic structures could be produced by allowing the sample to remain in the furnace for three minutes. Because a soaking period is desirable, the decision was reached that the time of austenitization should be five minutes.

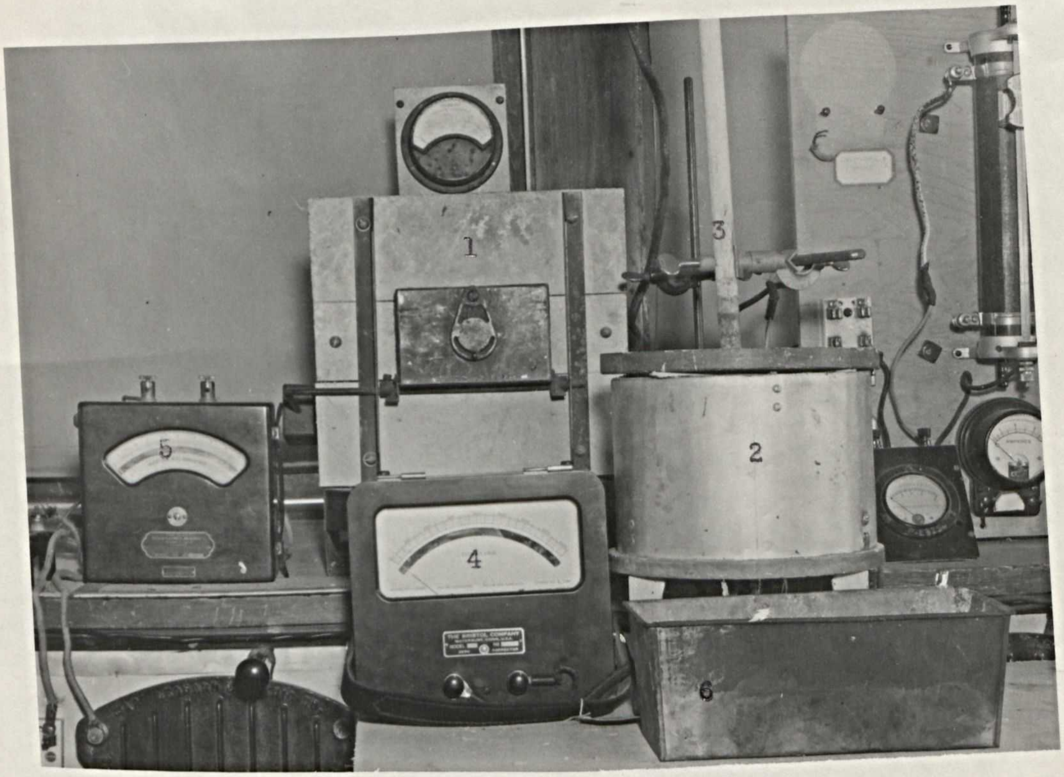
The isothermal tests were run in a manner similar to the previously described austenitizing time tests. Several small samples were placed

PLATE No. 4

Laboratory Equipment

PLATE No. 4

Laboratory Equipment



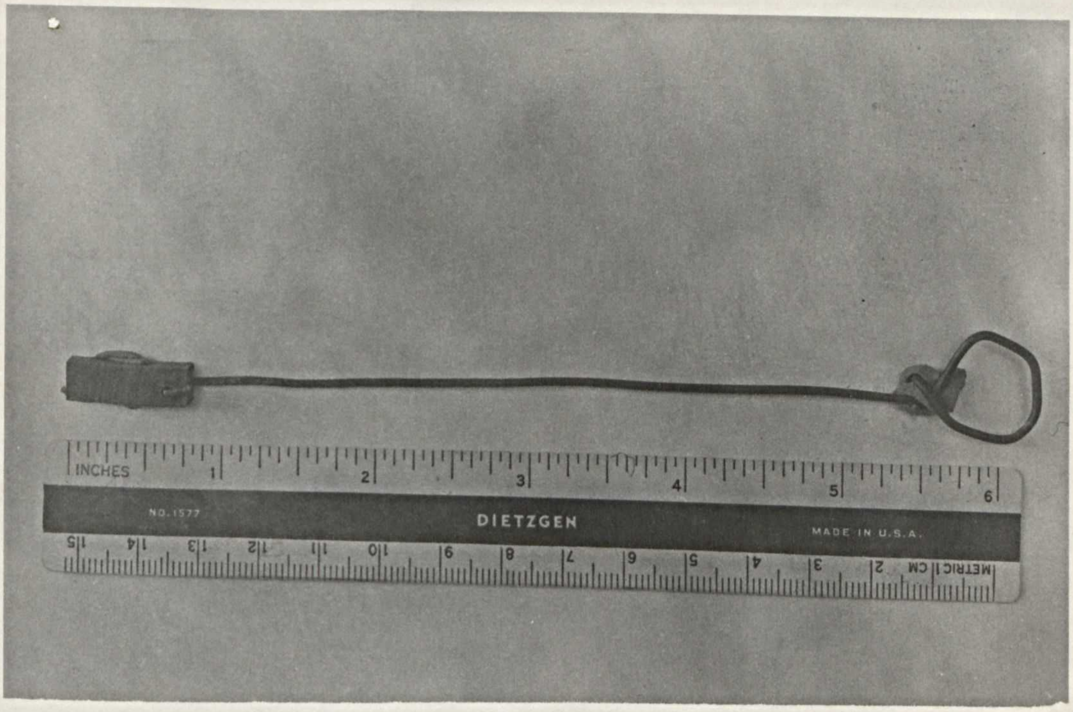
1. Austenitizing Furnace
2. Isothermal Transformation Furnace
3. Thermocouple
4. Bristol Pyrometer
5. Ammeter for Isothermal Furnace
6. Water Quenching Bath

PLATE No. 5.

Sample Holder.

PLATE No. 5

Sample Holder



in the austenitizing furnace whose temperature was 1650 F and allowed to remain in the furnace for five minutes. After this period, they were removed from the furnace and quenched into a bath whose temperature was in the range 1300 F to 600 F. The temperature of the bath was held constant in a pot type electric furnace as shown in plate $4. The exact temperature of the bath, which was molten lead, was recorded by the Bristol pyrometer shown in plate #4.

At various predetermined time intervals, a specimen was removed from the lead bath and quenched in the water bath. No strict schedule was employed for the time of removal from the lead bath, but rather the approximate time of start and of transformation was obtained from a general curve that was obtained from the literature (8) for a S.A.E. 6150 steel. Temperature intervals of 100 F from 1300 F to 700 F were investigated isothermally.

Hardness determinations on all the samples of the isothermal studies were made with the aid of a Rockwell Hardness instrument. An oxide scale was present on the samples, and because it gave erroneous results in the hardness determinations, it was removed by grinding on an 80-grit belt grinder. The results of the hardness determination will be reported in the next section of this work.

The metallographic inspection of the samples was conducted in the manner described by Viella (8) in the book, "The Metallographic Inspection of Steel." Very briefly this method consists of sectioning the samples on an automatic cut-off machine, mounting them in lucite molds; grinding them on an 80-grit belt grinder, and successive fine grinding om 1, 0, 00, 000, 0000, paperx; wet grinding on a wax lap with a suspension of 600-mesh alundum; polishing on a cloth lap with rouge; final polishing on a cloth lap with levigated aluminum oxide. The samples were etched with picral etching solution.

A conventional metallographic microscope was employed in the investigation on the microstructures. The reproductions of the microstructures that accompany this thesis were made with the aid of a Bausch and Lomb Metallograph.

RESULTS OF DETERMINATION

The results that were obtained by both Mr. Douglas and by the authors of this work are given in the following charts. A graph of the results is given in Plate #6. and, reproductions of the various microstructures appear in Plates 7--21 inclusive.

Metallographic Method

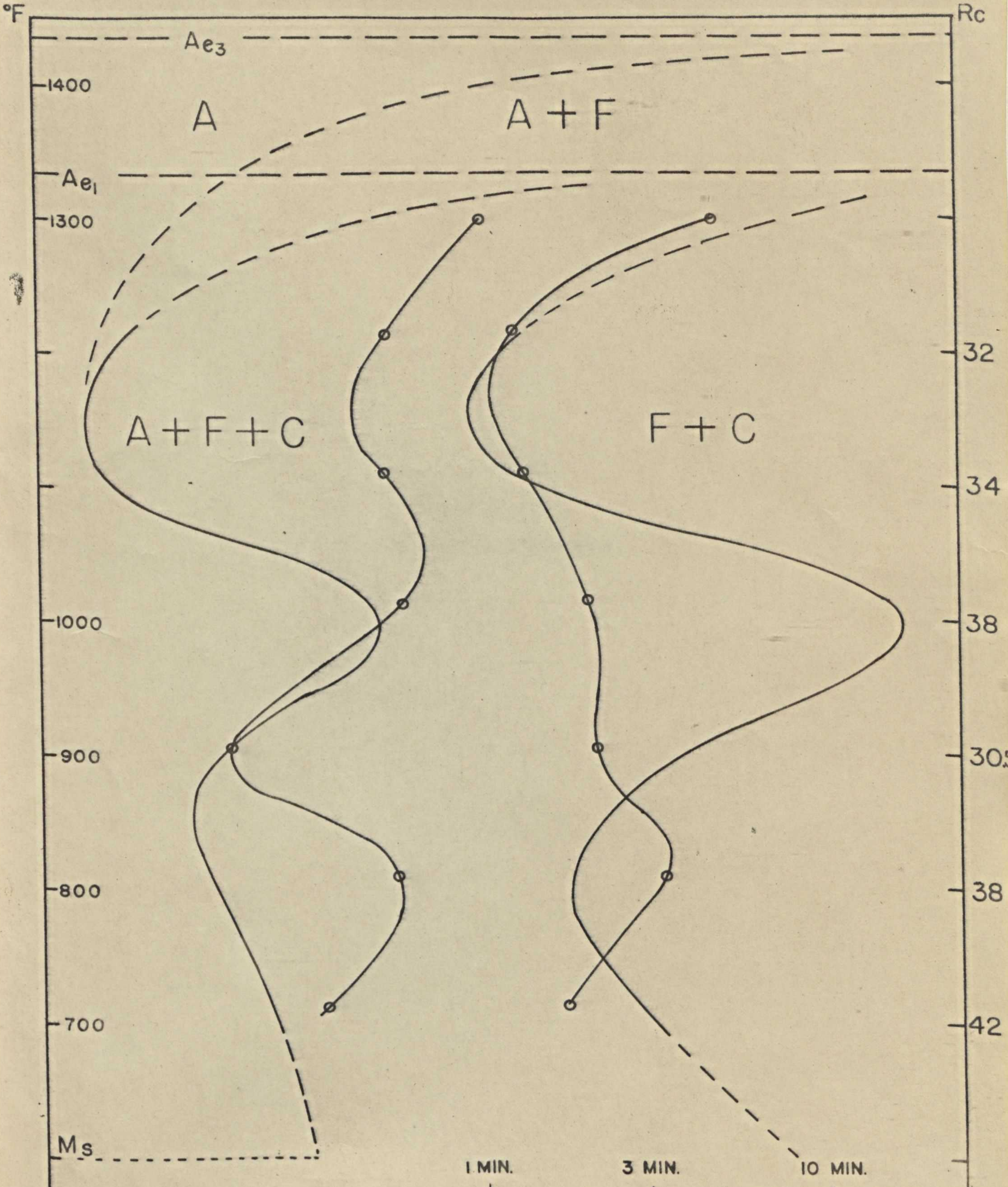
Temperature F	Time of Start of Transformation in seconds.	Time for Completion of Transformation in seconds	Hardness Rockwell C
1200	5	70	32
1100	30	90	34
1000	10	1020	38
900	10	210--240	30.5
800	10	90--120	38.0
700	15	180--210	42

Magnetic Permeability Method

1300	61.5	289	12
1200	31.1	72.4	20
1110	31.1	80.3	28
1015	34.5	122.4	28
905	10.8	127.1	28
810	33.9	204.8	28
715	20.5	106.1	28

PLATE No.6

6150 S-CURVE



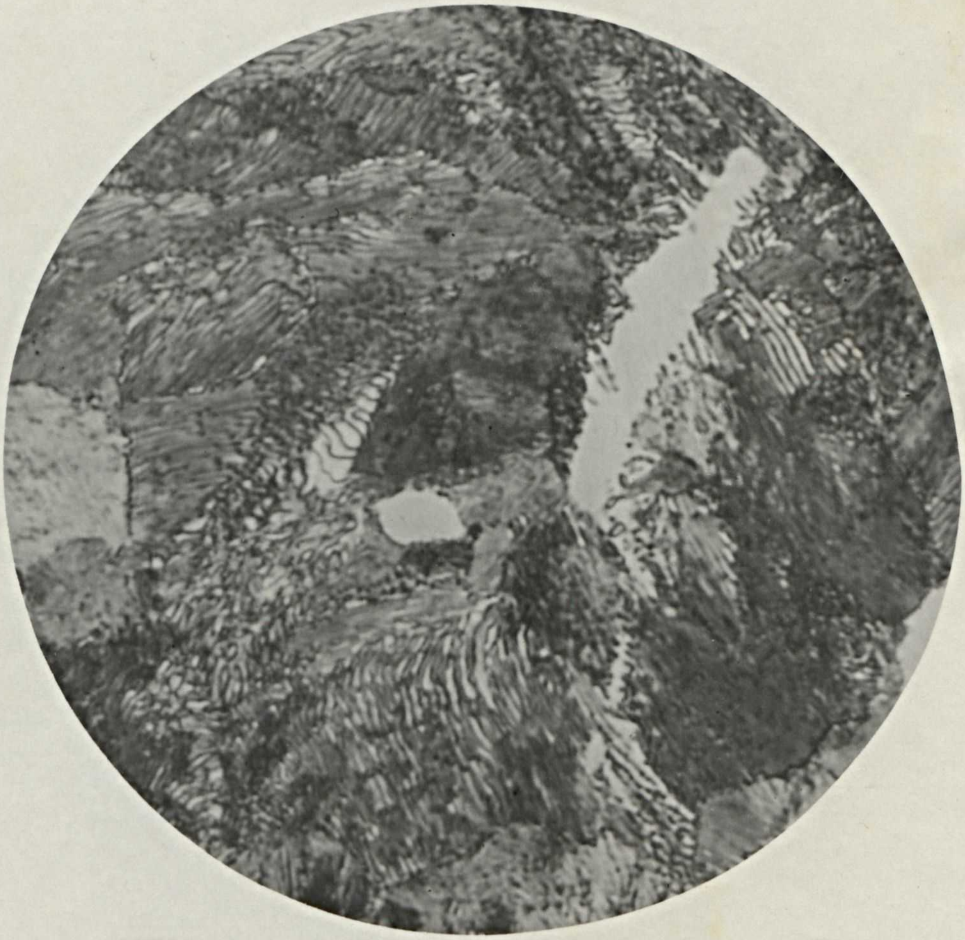
AUSTENITIZING TEMP. 1600°F

GRAIN SIZE NO. 7

PLATE No. 7

Normalized Structure

PLATE No. 7



Normalized Structure

Narrow, long inclusion of Manganese Sulfide and varying spaced lamellae of Ferrite and Cementite.

Mag. 785X. Enlarged 2X in reproduction

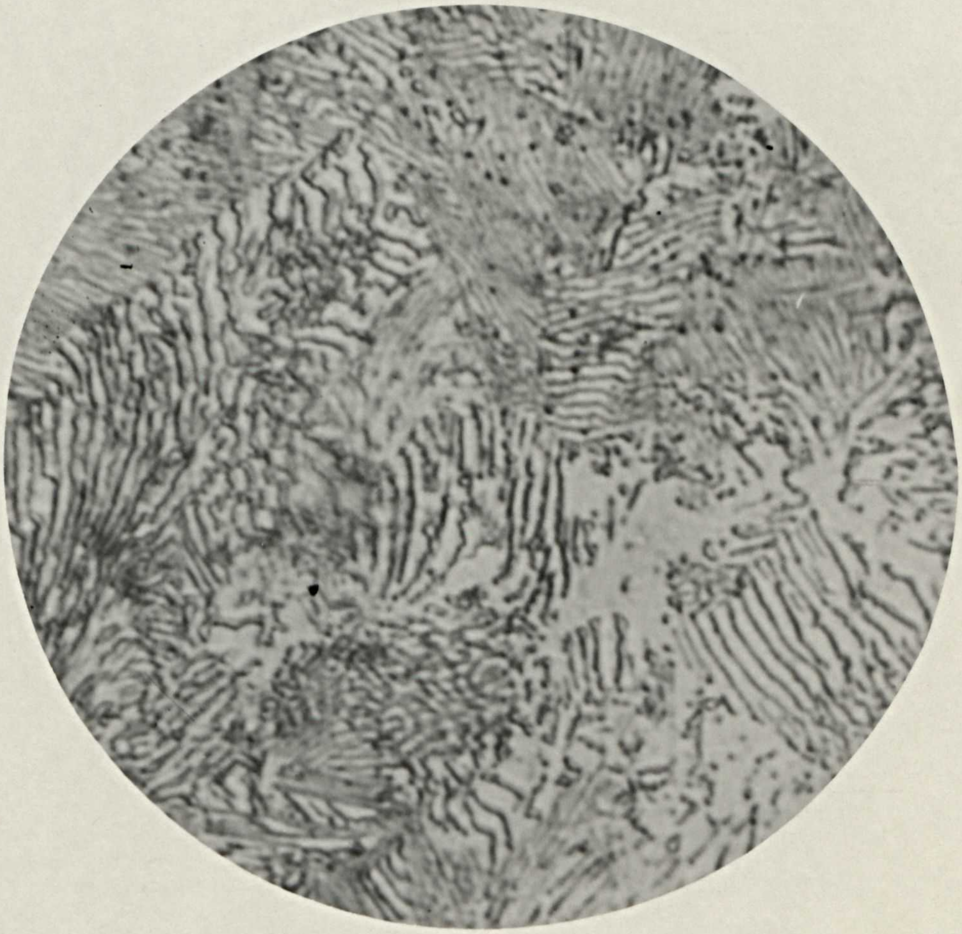
5% Picral Etch

PLATE No. 8

Annealed Structure

PLATE No.8

Annealed Structure



Uniform lamellaer pearlite

Mag. 785X. Enlarged 2X in reproduction.

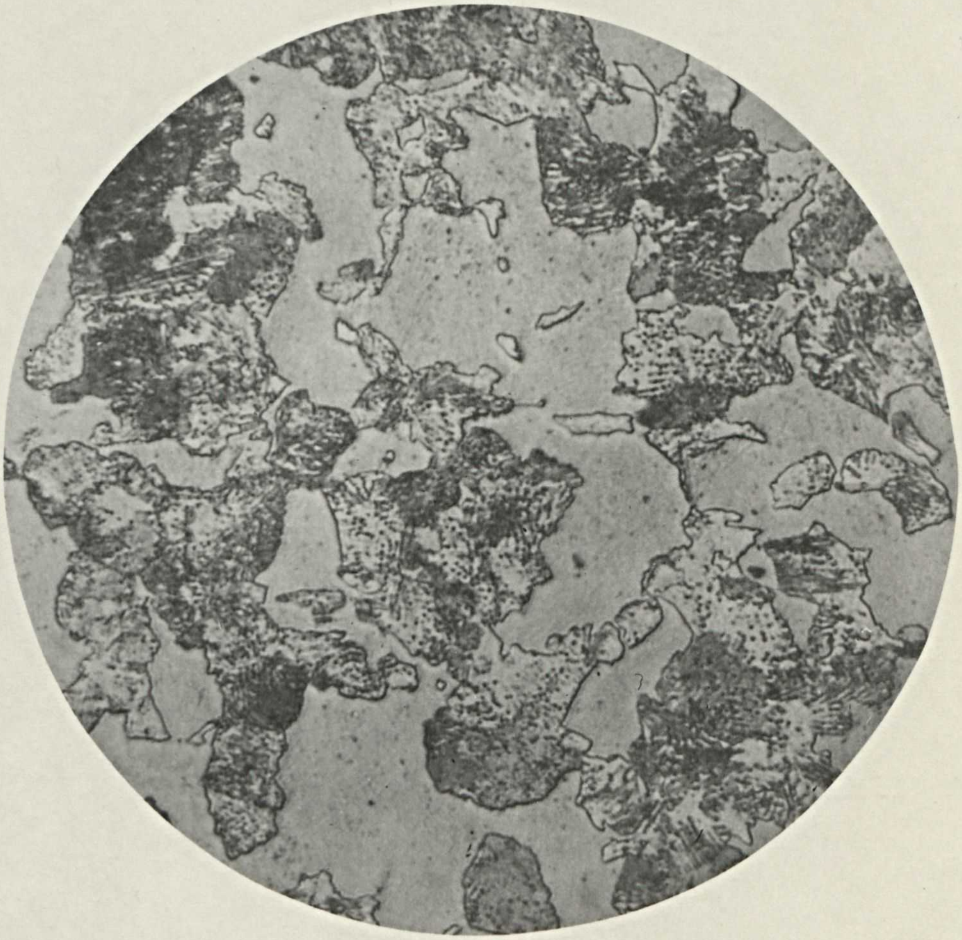
5% Picral Etch.

PLATE No. 9

Partial Transformation at 1200 F.

PLATE No. 9

Partial Transformation at 1200 F



Partially developed pearlite colonies in matrix of martensite.

Mag. 785X. Enlarged 2X in reproduction

5% Picral Etch.

PLATE No. 10

End of Transformation at 1200 F.

PLATE No. 10

End of Transformation at 1200 F

Structure primarily coarse pearlite.

Mag. 785X. Enlarged 2X in reproduction

5% Picral Etch.

PLATE No. 11

Near the Start of Transformation at 1100 F

PLATE No. 11

Near the start of transormation at 1100 F.

Small, nodular rosettes of pearlite in a matrix of martensite that was slightly tempered during grinding and polishing.

Mag. 785X. Enlarged 2X in reproduction.

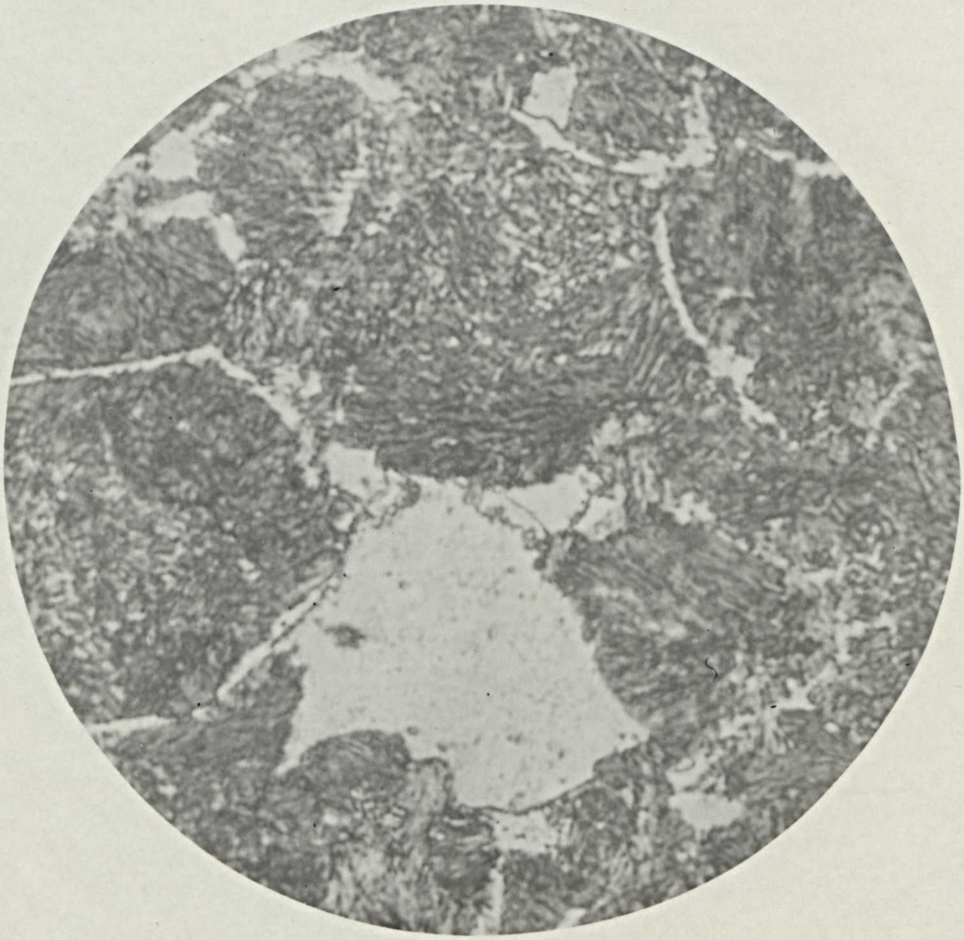
5% Picral Etch.

PLATE No. 12

End of Transformation at 1100 F.

PLATE No. 12

Transformation almost completed at 1100 F.



The small rosettes of pearlite shown in the previous plate have almost expended the austenite. The white area near the center is martensite.

Mag. 785X. Enlarged 2X in reproduction.

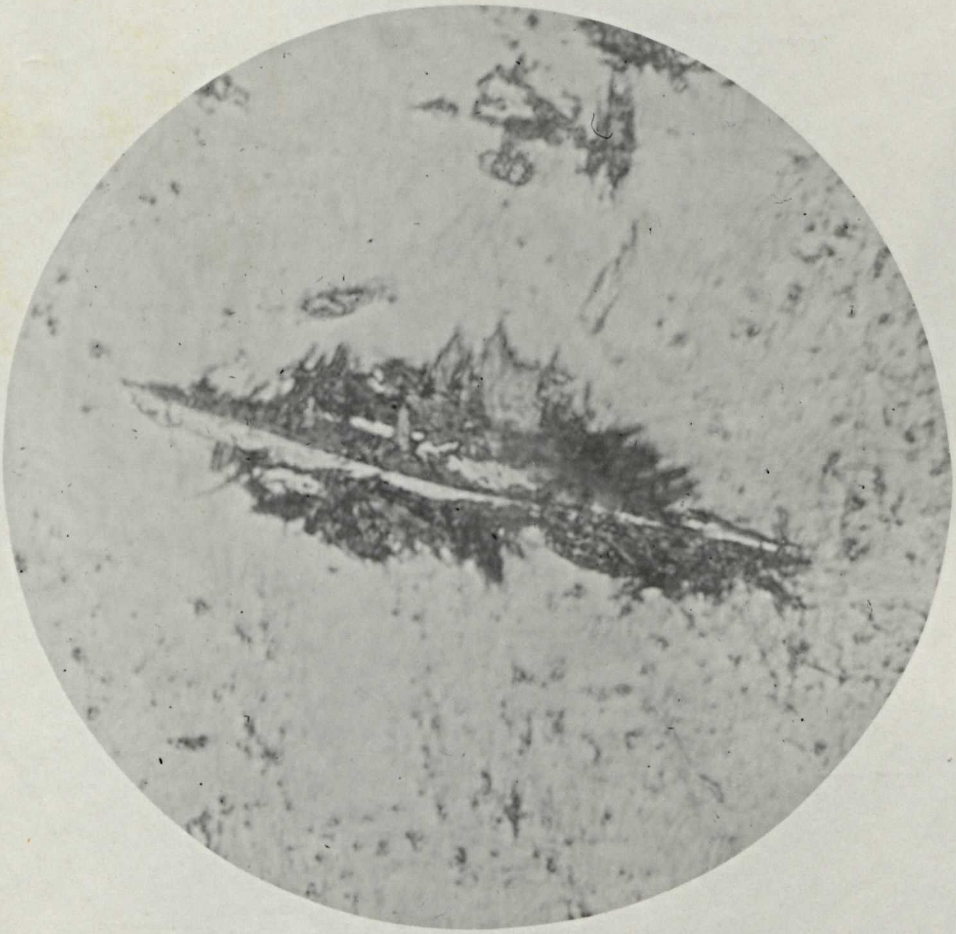
5% Picral Etch.

PLATE No. 13.

Start of Transformation at 1000 F

PLATE No. 13

Start of Transformation at 1000 F.



Bainite structure start to form at this temperature .

Mag. 785X. Enlarged 2X in reproduction.

5% Picral Etch.

PLATE No. 14

Partial Transformation at 1000 F.

PLATE No. 14

Partial Transformation at 1000 F.

Black transformation product is more acicular than that formed at higher temperatures. Matrix is martensite.

Mag. 785X. Enlarged 2X in reproduction.

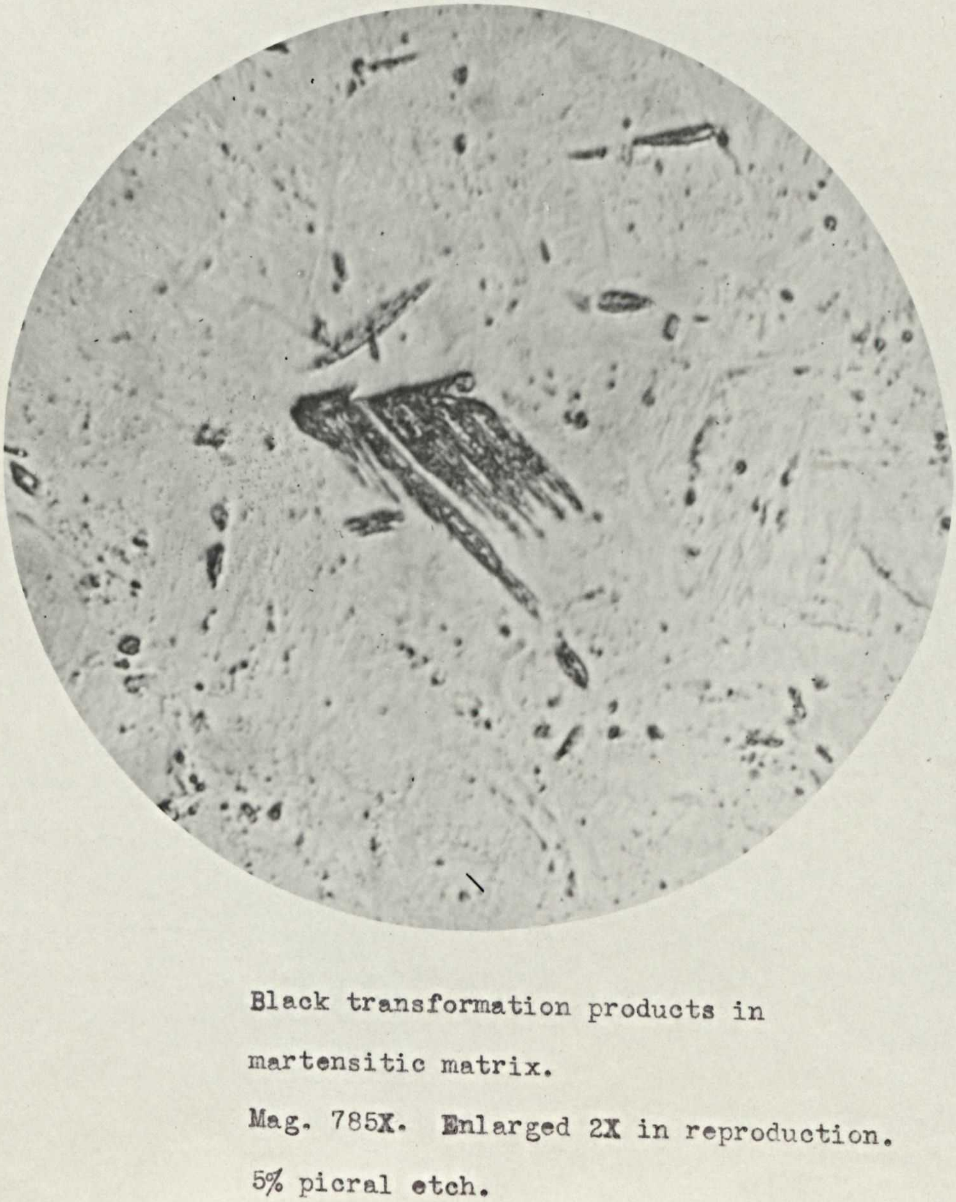
5% Picral etch.

PLATE No. 15.

Start of Transformation at 900 F.

PLATE No. 15

Start of Transformation at 900 F.



Black transformation products in martensitic matrix.

Mag. 785X. Enlarged 2X in reproduction.

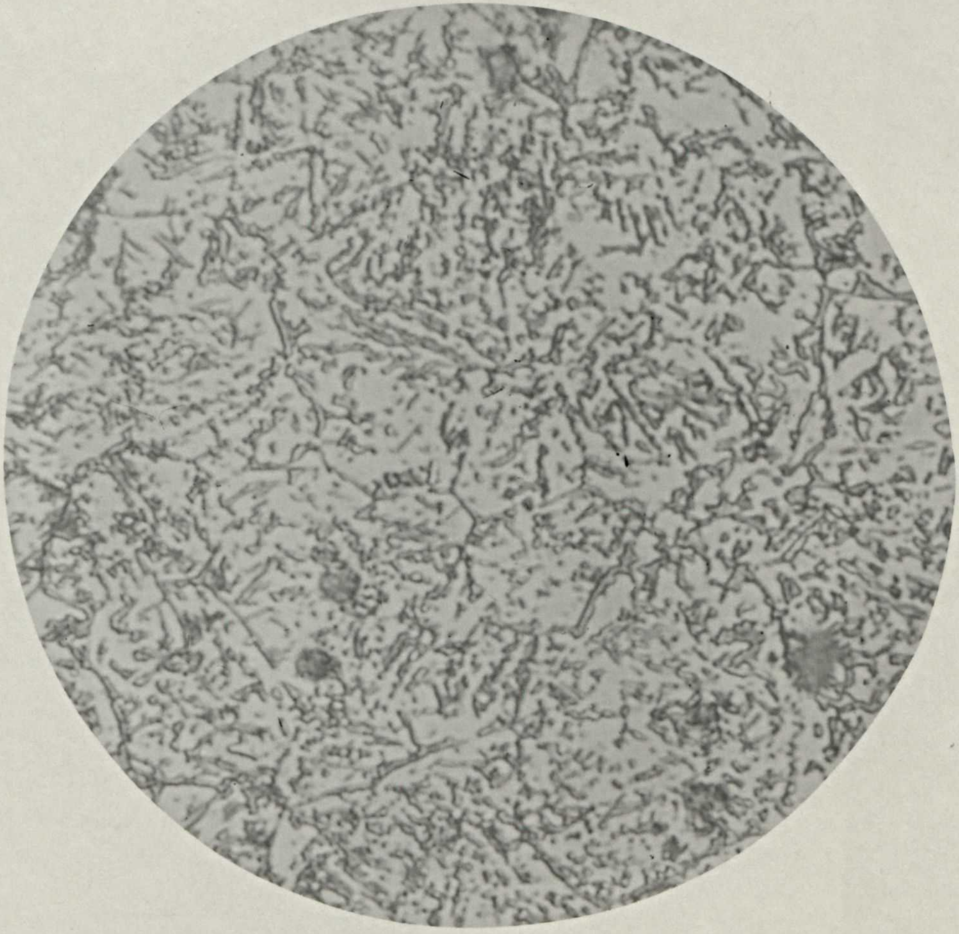
5% picral etch.

PLATE No. 16

Transformation at 900 F.

PLATE No. 16

Transformation at 900 F.



Sample held for an extended time after transformation was completed. Some spheridization is evident from microstructure.

Mag. 785X. Enlarged 2X in reproduction.

5% Picral Etch.

PLATE No. 17

Partial Transformation at 800 F.

PLATE No. 17

Partial Transformation at 800 F.

Black Bainite in a Martensitic Matrix.

Mag. 785X. Enlarged 2X in reproduction.

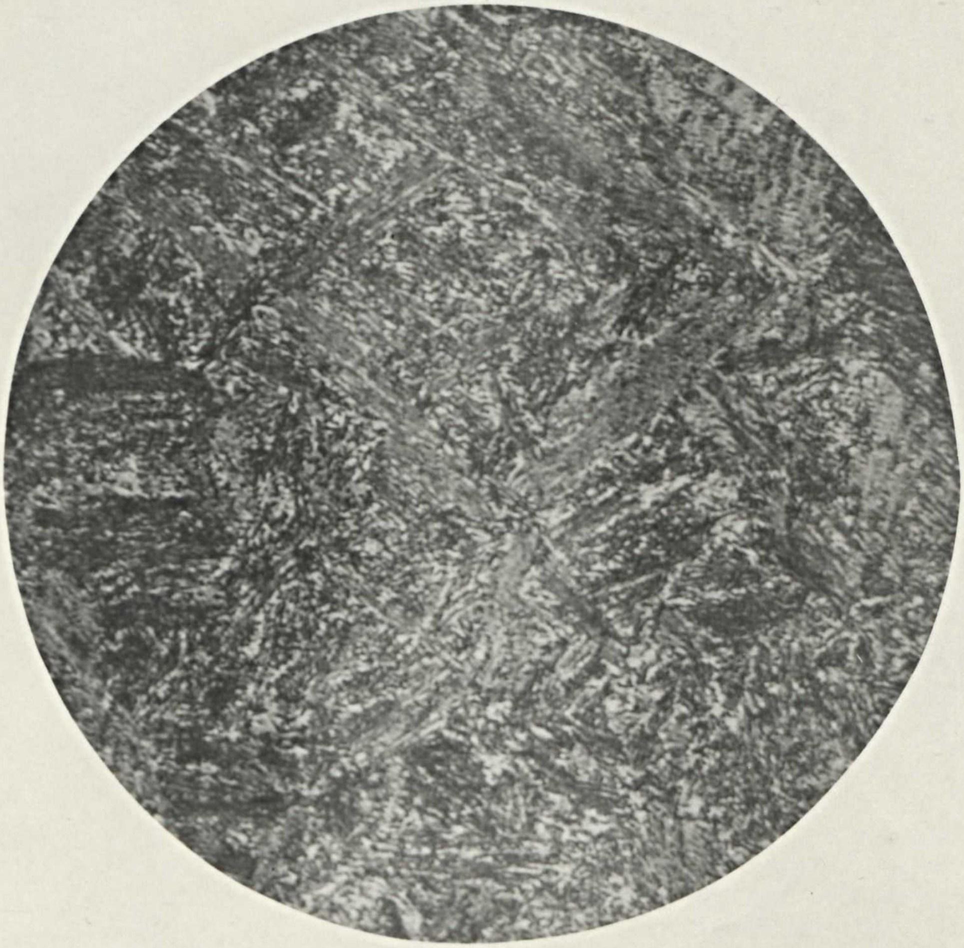
5% Picral Etch.

PLATE No. 18

End of Transformation at 800 F.

PLATE No. 18

End of Transformation at 800 F.



Black, aciciular Bainite transformation product that is difficult to resolve. Mag. 785X. Enlarged 2X in reproduction. 5% Picral Etch.

PLATE No. 19

Partial Transformation at 700 F.

PLATE No. 19

Partial Transformation at 700 F.

Black etching Bainite in white martensitic Matrix.

Mag. 785X. Enlarged 2X in reproduction.

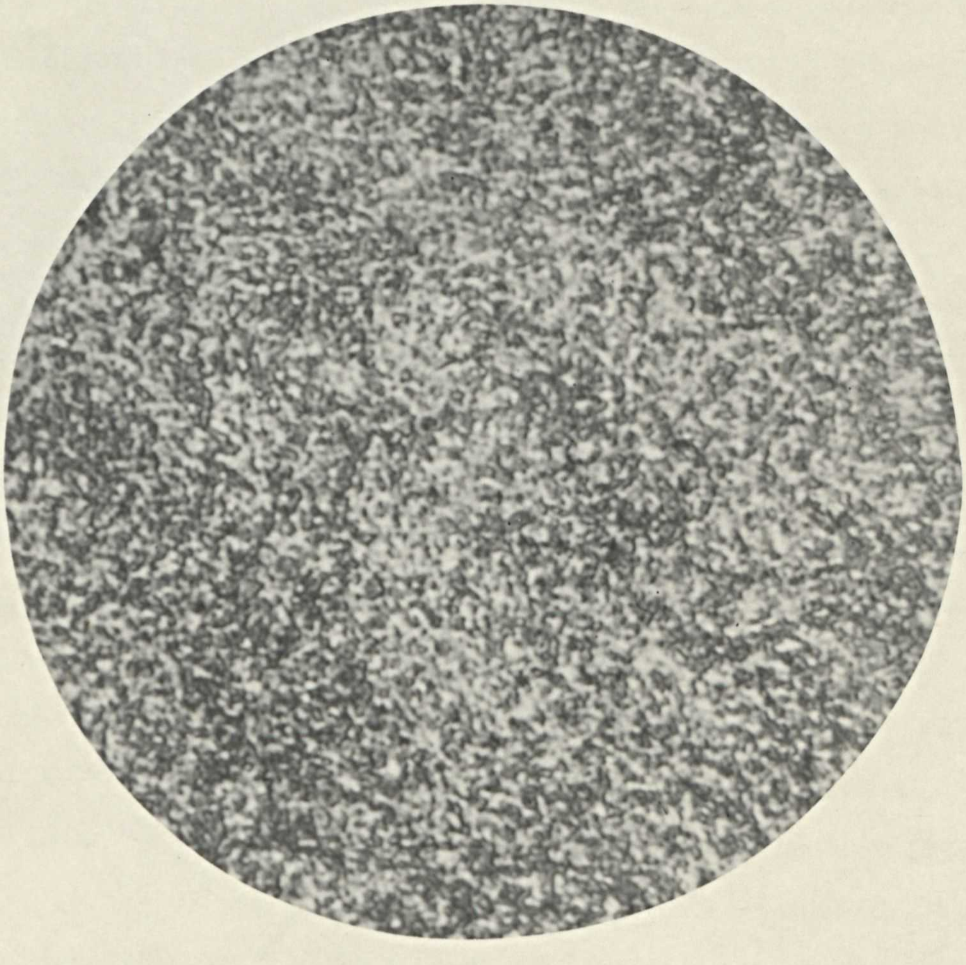
5% Picral Etch.

PLATE No. 20

End of Transformation at 700 F.

PLATE No. 20

End of Transformation at 700 F.



Fine carbides that are difficult to resolve.

Mag. 785X. Enlarged 2X in Reproduction.

5% Picral Etch.

DISCUSSION OF RESULTS

From the previous section and the graphical representation of the work, one can readily discern that there is very little agreement between the results that were obtained by the metallographical method with those that were obtained by the magnetic permeability method.

With the exception of two "Start of Transformation" points at 900 F and 1000 F, where one finds agreement, the magnetic permeability results lag the metallographic method. Down to 800 F, the general shape of the starting curve determined by each method is the same. The lagging effect noted in the magnetic permeability method could be due to flux losses in the magnetic circuit.

An austenite plus ferrite field is indicated by dotted lines for the metallographic determination. Dotted lines were chosen to represent the start of ferrite ejection. A structure that resembles ferrite in a martensite matrix was observed at 1300 F, 1200 F, 1100 F, and 1000 F, but the authers could not definitely distinguish it as proeutectoid ferrite, or retained austenite. In the sample that was chosen for the end of transformation at 1000 F, similar structures were not observed; consequently, this evidence would indicate that the structure that was observed as ferrite in the starting samples was in reality retained austenite. In the completed samples for 1100 F and 1200 F ferrite is present, but again the distribution of the ferrite in the microstructure is different than that noted in the starting samples.

A general curve for a S.A.E. 6150 steel, as given in the literature,(8) shows a similar ferrite field as is indicated in the graph of this work. In the curve from the literature, one notes that the time between ferrite ejection and pearlite ejection is very short--three seconds maximum; therefore, the exact determination of a ferrite field would be difficult to achieve with the equipment that was employed in this determination.

At 1100 F and 1200 F the results of the end of transformation are in good agreement, and at 1300 F the results may be in agreement for at this temperature the curve flattens out and a slight variance in temperature would give a wide variance in time. The results of the metallographic determination indicated that an infinite time period would be needed at this temperature; whereas, the magnetic permeability method gave approximately four minutes. No definite criticism of either result should be made because of the afore mentioned trend of the curves.

Below 1100 F the curves will fail to agree. The results that were obtained for the magnetic permeability method from 1100 F to 700 F do not vary; moreover, the hardness measurements do not increase. In comparison to this, the results that were obtained by the metallographic method show a lengthy period of transformation at 1000 F. The curve then turns back to the shorter time period down to 800 F. At 700 F the curve is directed again to the longer time periods. The hardness values that were obtained in the metallographic determination are more in agreement with expected values for the type of steel under investigation.

The curve from the literature (8) resembles the curve that was obtained from the metallographic determination. The only difference is that it is displaced approximately 100 F downward; e.g.; the long period for time of transformation comes at 900 F instead of 1000 F as is shown in this work.

The results of a microscopic investigation of the samples that Mr. Douglas used for his determination showed that the structures that were obtained for the points in question did not correspond to those obtained by the metallographic determination. They resembled the structure that was formed at 1200 F. (This is shown in plate #10). A structure such as this would agree with the hardness results that Mr. Douglas obtained, but it would not be in agreement with the structures that should form at the lower temperature levels.

To obtain the pearlitic structure as shown in Mr. Douglas' samples, one would have to provide normalizing or annealing conditions. Therefore, the results that were obtained in the magnetic work are erroneous, not because of the failure of the magnetic circuit, but rather because of transformation taking place at a higher temperature than that desired.

One can conjure two general conditions that may have influenced the type of a structure that was produced. If the temperature of the furnace that was employed to bring about the gamma condition in the samples was below the temperature required, but sufficiently high enough for annealing, the structure would be formed. However, if this happened the time of start of transformation would be instantaneous, and since the results that were presented actually lag those that were obtained metallographically, one is led to discard this reasoning.

Another, and perhaps the more logical explanation, would concern the quenching furnaces that were used in the determination. A full description of the construction of the furnace is given in a thesis that was prepared by Mr. Joshi.(9) The reader is referred to it for a complete discussion of the construction of the furnace and of the magnetic circuit used in the tests. If this furnace was at a higher temperature than that indicated in the results, the normalized structure would result. The author of this work is inclined to believe that this may well be the reason for the discrepancies in the curves for the end of transformation. As was previously mentioned, flux losses are probably the cause of the lagging of the curve that indicates start of transformation.

CONCLUSIONS

Because of the wide discrepencies between the results obtained by the two methods, and in light of the reasoning as to the cause of the discrepencies that were presented in the previous section, no attempt will be made to present any conclusions to the relative merits of the two methods. The author believes that the curve that is presented for the metallographic results is accurate within the limits of the equipment that was employed. Until a more extensive test of the magnetic permeability method is made, one should reserve judgement as to its applicability. Undoubtedly, if successful it would offer a very attractive method of study. The work of preparing S-curves could be greatly reduced.

Considering the curve as determined by the metallographic method as representative of the steel under test, a discussion of a few methods of heat treating practices as regards to time and temperature can be given, and some general conclusions can be drawn. They are as follows:

1. If cyclic annealing is to be used on this steel, the temperature should be approximately 1200 F. This would give the most economical point of operation, because temperatures near 1300 F would require infinite time; a like condition is found near 1000 F.

2. To produce a 100 per cent martensitic structure by quenching from the gamma range, a drastic quenching medium will be necessary. If slow quenching methods are used, transformation will occur at the higher temperatures and in the range of 900 F, because of the rapidity for the start of transformation at those temperatures.

3. In comparison to a plain carbon steel this steel is sluggish to finish transformation once transformation has started. This is characteristic of all chromium steels.(3)

RECOMMENDATIONS

The desirability of repeating the magnetic method as a check for this determination has already been stated. If this is done, the beginning and end points of the curve that were obtained by the metallographic method should be checked by a metallographic determination.

Determinations at temperatures above 1200 F should be repeated if the curve is to be considered complete. Very little work was done in this region, and therefore, the results may be slightly in error. Moreover, the M_s and M_f temperatures of the steel should be determined by experiment. The values that are given on the graph are taken from the literature (8) for a general 6150 steel; therefore, they may be in error.

Finally, if similar metallographic determinations are to be conducted at this institution, one would find it attractive if an electrolytic polishing unit was available. Also the metallograph that was used for the reproduction of the microstructure is not the most convenient instrument one could employ. If a newer unit was available the work would be expediated.

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Since this was a joint problem, this author wishes at this point to record some tangible evidence of his thanks to the other member, Mr. C. Arentzen. He proved a most intelligent and diligent worker.

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