

A Thesis
on
The Mechanism of Solidification of Metals.

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by

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SUMMARY OF CONTENTS.

List of Illustrations	IV.
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Introduction	1.
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SECTION ONE.

Previous Work

1. General Review of Literature associated with the Mechanism of Solidification	5.
2. More Recent Investigation, primarily associated with the Solidification of Sand Castings	29.

SECTION TWO.

A Metallurgical Investigation into the Effect of Variation in the Method of Feeding on the Solidification of Two Series of Sand Cast Steel Blocks.

Synopsis	37.
Definition of Terms	38.
Series I - Experimental Details	38.
Results: Structure of Castings	43.
Chemical Analysis	50.
Mechanical Properties	51.
Series II - Experimental Details	53.
Results	54.
Preliminary Comments on the Results of the Examination of the Two Series of Industrial Castings	60.

SECTION THREE.

Introduction	63.
Experimental Techniques	
(a) Design of Castings	65.

SECTION THREE (Contd).

Experimental Techniques

(b) Moulding Technique	66.
(c) Mould Assembly	68.
(d) Temperature Measurement	70.
(i) Differential Temp. Measurement	71.
(ii) Direct Temp. Measurement	74.
(e) Modification to the Moulding Technique necessitated by Temperature Determinations	80.
(f) Modification to Mould Arrangement for "Pour-Out" Tests	85.

Melting and Casting Procedures

(a) The Effect of Size Factor	87.
(b) The Effect of Casting Temperature	90.
(c) Temperature Measurements	
(i) Differential Method	91.
(ii) Direct Method	92.
(d) "Pour-Out" Tests	94.

Examination of Castings 96.

Results (a) Size Effect	97.
(b) Casting Temperature	105.
(c) Temperature Measurements	
(1) Differential Method	109.
(2) Direct Method	117.
(d) "Pour-Out" Tests	128.

SECTION FOUR.

Discussion	129.
1. Introduction	129.

SECTION FOUR (Contd.)

2. Superheat and Casting Temperature	132.
3. Convection Currents and Mould Arrangement	134.
4. Size Effect	139.
5. The Mechanism of Solidification	
(a) Theories of Solidification	145.
(b) Consideration of the Results of the Temperature Measurements obtained during the Present Investigation in Relation to the Solidification Process	155.
(c) "Pour-Out" Experiments	169.
6. Interface Temperature & Superheat	170.
(i) Interface Temperature	171.
(ii) Superheat	174.
7. Falling Crystallite Theory	177.
8. Microscopical Examination	184.
9. Summary of Conclusions	187.
References	194.
Appendix I Sulphur Printing	199.
Appendix II Etching Technique	200.

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List of Illustrations.

1.	Arrangement of Castings.	between p.37 and 38
2.	Structure of Casting Al.	" p.60 " 61
3.	Macrostructure of Casting Al.	" " " "
4.	Sulphur Print of Casting Al.	" " " "
5.	Structure of Casting Bl.	" " " "
6.	Macrostructure of Casting Bl.	" " " "
7.	Sulphur Print of Casting Bl.	" " " "
8.	Structure of Casting Cl.	" " " "
9.	Macrostructure of Casting Cl.	" " " "
10.	Sulphur Print of Casting Cl.	" " " "
11.	Structure of Casting Dl.	" " " "
12.	Macrostructure of Casting Dl.	" " " "
13.	" " " "	" " " "
14.	Sulphur Print of Casting Dl.	" " " "
15.	Structure of Casting El.	" " " "
16.	Macrostructure of Casting El.	" " " "
17.	" " " "	" " " "
18.	Sulphur Print of Casting El.	" " " "
	TABLE I Chemical Analysis Casting Al.	" " " "
	TABLE II " " " Bl.	" " " "
	TABLE III " " " Cl.	" " " "
	TABLE IV " " " Dl.	" " " "
	TABLE V " " " El.	" " " "
	TABLE VI Mechanical Properties" Al.	" " " "
	TABLE VII " " " El.	" " " "

19.	Structure of Casting A2.	between p.60 and 61.
20.	Macrostructure of Casting A2.	" p.69 " 70.
21.	Sulphur Print of Casting A2.	" " " "
22.	Structure of Casting B2.	" " " "
23.	Macrostructure of Casting B2.	" " " "
24.	Sulphur Print of Casting B2.	" " " "
25.	Structure of Casting C2.	" " " "
26.	Macrostructure of Casting C1	" " " "
27.	Sulphur Print of Casting C2.	" " " "
28.	Structure of Casting D2.	" " " "
29.	Macrostructure of Casting D2.	" " " "
30.	Sulphur Print of Casting D2.	" " " "
31.	Structure of Casting E1.	" " " "
32.	Macrostructure of Casting E2.	" " " "
33.	Sulphur Print of Casting E2.	" " " "
34.	Pattern A.	" p.65 " 66.
35.	Pattern B.	"
36.	Pattern C.	"
37.	Pattern D.	"
38.	Pattern E.	"
39.	View of Open Half Mould with Exothermic Cores	p. 68
40.	Top View of Open Mould showing the Head and position of Thermocouple Sheaths	facing p.69.
41.	Side View of Mould Assembly showing Thermocouple Connections	between p.69 and 70.
42.	End View of Mould Assembly showing Thermocouple Connections	" " " "
43.	General View of Mould Assembly & Furnace	" " " "

44.	Diagrammatic Circuit of Recording Apparatus.	between p.70 and 71
45.	Thermocouple Connections for Difference Measurements.	p.72.
46.	Section of Recorder Chart showing typical pen trace for Temperature Difference Measurements.	p.75.
47.	Section of Recorder Chart showing typical pen trace for Direct Temperature Measurements.	p.77.
48.	Thermocouple Sheath Assembly.	p.81.
49.	Guards for Thermocouple Sheaths.	p.84.
50.	" " " "	p.84.
51.	Mould Inversion Arrangement for Pour Out Tests.	facing p.86.
52.	Structure of Casting A. WOX6 - SHA4.	between p.104 & 105.
53.	" " " B. WOX7	" " " "
54.	" " " C. WOX3	" " " "
55.	" " " D. WOX8, WX2	" " " "
56.	" " " E. WOX5, WX1, WX3.	" " " "
57.	Macrostructure of Casting A. WOX6.	" " " "
58.	" " " B. WOX7.	" " " "
59.	" " " C. WOX3.	" " " "
60.	" " " D. WX2.	" " " "
61.	" " " E. WX3.	" " " "
62.	Structure of Casting SHA1.	" p.108 " 109
63.	" " " SHA2.	" " " "
64.	" " " SHA3.	" " " "
65.	" " " SHA5.	" " " "
66.	" " " SHA6.	" " " "

67. Macrostructure of Casting SHA1. between p.108 and 109.
68. " " " SHA2. " " " "
69. " " " SHA3. " " " "
70. " " " SHA5. " " " "
71. " " " SHA6. " " " "
72. Temperature Difference Curves obtained
between positions on Central Axis of
Top-Fed Vertical Block - Casting T61A. facing p.110.
73. Temperature Difference Curves obtained
between positions on Central Axis of
Bottom-Fed Vertical Block - Casting T62A. " p.113.
74. Temperature Difference Curves obtained
between positions on Transverse Section
of Top-Fed Vertical Block - Casting T61E. " p.114.
75. Temperature Difference Curves obtained
between positions on Transverse Section
of Bottom Fed Vertical Block - Casting T62E. " p.116.
76. Iron - Carbon Equilibrium Diagram. " p.116.
77. Cooling Curves obtained at positions on
Transverse Section of Top-Fed Vertical
Block - Casting A1. " p.120.
78. Cooling Curves obtained at positions on
Transverse Section of Top-Fed Vertical
Block - Casting A2. " p.121.
79. Cooling Curves obtained at positions on
Central Axis of Top-Fed Vertical Block
Casting A3. " p.122.
80. Cooling Curves obtained at positions on
Transverse Section of Bottom Fed Vertical
Block - Casting E1. " p.124.
81. Cooling Curves obtained at positions on
Transverse Section of Bottom Fed Vertical
Block - Casting E2. " p.125.
82. Cooling Curves obtained at positions on
Central Axis of Bottom Fed Vertical
Block - Casting E3. " p.126.

83.	Casting Bled After One Minute.	Between p.129 and 130.
84.	" " " Two Minutes.	" " " "
85.	" " " Three "	" " " "
86.	" " " Four "	" " " "
87.	Effect of Size of Feeder-Head on Macrostructure of Top-Fed Vertical Sand Castings (5" x 3" x 2½")	" p.141 and 142.
88.	Effect of Size of Feeder-Head on Macrostructure of Top-Fed Vertical Sand Castings (12" x 6" x 4")	" p.141 and 142.
89.	Temperature Distribution in Liquid and Solid Steel.	page. 149.
90.	Variation of liquidus temperature with distance from Mould Wall, at time t.	page. 151.
91.	Comparison of Solidification Curves for Chill and Green Sand Castings.	page. 155.
92.	Curves representing the Process of Solidification across Top-Fed Block (0 - 4 mins)	Between p.167 and 168.
93.	Curves representing the Process of Solidification across Top-Fed Block (4 - 7 mins)	" " " "
94.	Curves representing the Process of Solidification across Bottom Fed Block (0 - 4 mins)	" " " "
95.	Curves representing the Process of Solidification across Bottom Fed Block (4 - 6 mins)	" " " "
96.	Solidification Curves of Castings A1 and A2	" p.167 " 168.
97.	Solidification Curves of Castings E1 and E2	" p.167 " 168
98.	Nuclear Crystals Casting A1 x12	" p.186 " 187.
99.	Nuclear Crystals Casting A1 x35	" p.186 " 187.
100.	Chill Crystals Casting E1 x35	" p.186 " 187.

- | | | | |
|------|--------------------------------|------|---------------------|
| 101. | Columnar Crystals. Casting E1 | x35. | between p.186 & 187 |
| 102. | Equiaxial Crystals. Casting E1 | x35. | " p.186 & 187. |
| 103. | Nuclear Crystals. Casting E2 | x12. | " p.186 & 187. |
| 104. | Nuclear Crystals. Casting E2 | x35. | " p.186 & 187. |
| 105. | Dendritic Crystals. Casting E2 | x12. | " p.186 & 187. |
| 106. | Dendritic Crystals. Casting E2 | x35. | " p.186 & 187. |

SECTION ONE

INTRODUCTION.

Object.

The object of the research was to investigate in detail some of the factors influencing the mechanism of solidification of sand cast steel blocks, with particular reference to the effect of variation in the method of feeding and in the casting temperature.

The results of many previous investigations have shown the complex nature of the effects of various independent variables associated with the solidification process, particularly in the case of ingots in metal moulds. However, as relatively little is known of the nature of solidification of sand castings, it is hoped in the present study to establish a satisfactory theory of the effect of certain variables in the mechanism of solidification of such castings.

In past investigations the effects of such variables as composition, pouring temperature, mould design, etc., on the solidification process have been studied by techniques that can be divided into two main groups:-

- A. Examination of the solidified ingot.
- B. Examination of the conditions existing during the solidification period.

By an examination of the solidified ingot the resultant effects of the different variables could be ascertained, whereas by determining the conditions which existed during the solidification process information of a more fundamental nature could be obtained. The first method generally consisted of microscopical and macroscopical examination, chemical analysis and mechanical testing. The second necessitated an entirely different approach, which has been made in four ways:-

- (1) Mathematical analysis; (2) "Pour-out" methods;
- (3) Electrical "Analogue"; (4) Direct temperature measurement.

The mathematical approach has had some success in indicating the form of the temperature distribution, but it can only be applied to simple cases and cannot be extended to treat all the complex variables associated with the process. Furthermore, data concerning the thermal properties of materials at high temperature are scarce and liable to be inaccurate.

"Pour-out" experiments show the approximate extent of solidification after different times, but give little information about the thermal conditions or the nature of the "unsolidified" metal. For example, with alloys possessing a freezing range, the solidified shell always retains liquid material in the crystal interstices, which solidifies after the "pour-out" has been made.

In the electrical analogue method a comparison is made between an electrical circuit and a set of heat conditions. This method can be informative but like the mathematical approach it is handicapped by lack of knowledge of the thermal properties of materials. It cannot be applied to study conditions between the liquidus and solidus temperatures of alloys.

Direct temperature measurement is the most satisfactory method, as it enables a continuous record to be made of the actual conditions existing during the solidification process. It also possesses a number of distinct advantages over the methods described above. For instance, it does not require a knowledge of the thermal properties of materials, and it can be employed to study the conditions between the liquidus and solidus temperatures. Mainly because of the practical difficulties involved in the technique, however, it has received relatively little attention.

All these methods are reviewed in greater detail in Section I of the thesis.

In the present investigation a preliminary study is made of the effect of variation in the method of feeding of sand cast steel blocks; this has been enlarged to include the determination of the effect of size factor and casting temperature.

A series of direct temperature measurements has been made, in conjunction with "pour-out" experiments, in an attempt to relate the different types of crystallization with the actual temperature conditions.

These investigations are described in detail in the thesis in two main sections. In Section I the effect of the variation of feeding in the structure of two series of sand cast steel blocks, manufactured under industrial practise is described. In Section II a description of the effect of size factor, casting temperature and measurement of the actual temperature conditions is given.

Finally, the various aspects associated with solidification are initially discussed and a probable theory advanced to explain certain aspects of the mechanism of solidification, in the light of the results obtained from the investigation.

PREVIOUS WORK.

1. GENERAL REVIEW OF LITERATURE ASSOCIATED WITH THE MECHANISM OF SOLIDIFICATION.

Although much of the published work has been concerned with the study of the mechanism of solidification of steel ingots in metal moulds, many of the principles that have been deduced apply in general to the solidification of sand castings. However, there are numerous inconsistencies, and various phenomena still remain unexplained. For example, much attention has been devoted to the interpretation of the transition that occurs between the columnar and equiaxial growth frequently found in cast metal, and yet no satisfactory theory has been proposed to explain all the cases in which these two structures occur.

Desch¹ and Belaiew² surveyed and discussed the early literature on the subject of crystallization, covering the demonstration of the crystalline nature of metals by Grignon, Sorby, Tchernoff and others, and speculation regarding the mechanism of nucleation and dendritic growth by such workers as Tammann, Lehmann, Ostwald and Miers.

Tammann³ considered that the nucleation and growth of organic compounds are governed by two principal laws, (a) that the number of nuclei formed increases to a maximum as the degree of undercooling of the liquid increases, decreasing beyond this to zero for still higher degrees of undercooling, and (b) that the velocity of crystallization

increases rapidly with increasing undercooling, but finally attains approximately constant value.

Although Tammann refuted⁽¹⁾ the theoretical conclusions of Ostwald and Miers that there are two stages of undercooling, viz. the metastable and labile states, Desch⁴ has stated that these conditions result in the columnar and equiaxial types of crystallization respectively. The metastable condition is reached when a liquid has cooled to such a temperature that it is incapable of crystallizing spontaneously but will crystallize if inoculated with the solid phase; the labile state is attained when, on further cooling, the liquid is able to crystallize spontaneously without inoculation.

The mechanism of formation of crystal dendrites was originally discussed by Lehmann¹ in terms of the crystallization and diffusion of salt solutions, but his conclusions also apply to the formation of metal dendrites. Once a crystal nucleus has formed in a salt solution, growth will continue, according to Brevais' Law, preferentially along certain axes. If the crystal is allowed to grow slowly, it develops into a normal idiomorphic crystal. However, under conditions of rapid growth, the corners of the crystallite act as protuberances which extend into the solution. In the immediate vicinity of the growing crystal, concentration gradients are set up and the continuation of growth is dependent on the diffusion of material from the

bulk of the solution into the depleted region in contact with the crystal. Simple geometric considerations show that the solution is replenished most rapidly at corners, with the result that growth is accentuated along the crystallographic axes. By similar reasoning, it was shown how secondary and tertiary branches could be formed by this process.

As diffusion of heat and the diffusion of matter obey similar laws, the above explanation accounts for the growth of pure metal dendrites from the liquid state. In the same way, the formation of dendrites in a solidifying alloy can be understood by considering both the thermal gradients and concentration gradients in the system. This is discussed later.

One of the first investigations of the factors influencing the solidification of industrial ingots was made by Howe⁵ in 1908. His conclusions on the causes and "the restraining" of piping and segregation were of fundamental interest and indicated the direction in which further research was required. He concluded, and later demonstrated, in some cases with wax ingots, that the pipe and segregate could be controlled by means of sink heads and by controlling (a) the rate of pouring (b) the pouring temperature (c) the condition of the metal (whether killed or not) and (d) the rate of cooling. In discussing segregation, he introduced the principle of differential solidification to explain the

formation of "impurity" rich areas. Although certain of his conclusions concerning the mode of crystal formation appear unlikely, the ideas that he advanced are the basis of many theories presented in more recent times.

In a latter paper (1916),⁽⁶⁾ Howe explained the transition between columnar and equiaxial crystallization on the basis of temperature gradients, composition gradients and rates of cooling. This explanation is basically the same as that advanced by Northcott⁷ in 1941, although the latter developed it along more scientific lines.

With many of the fundamental aspects of solidification and crystallization satisfactorily explained, the scope of later investigations became more closely defined, being concerned more with the determination of specific features associated with the mechanism of solidification. For example, we may consider the work of the Heterogeneity of Steel Ingots Committee⁸, which included the following:-

- (a) Mathematical analysis of thermal conditions.
 - (b) Direct quantitative examination of solidification rates.
 - (c) Observation of the factors influencing primary crystallization.
 - (d) Determination of physical data.
 - (e) Electrical analogue experiments.
 - (f) Thermodynamical considerations.
- (a) Mathematical Investigation.

⁹
Ruddle, in a review of the literature, summarized the important methods applied to the study of solidification.
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He mentioned the early attempt of Saito⁽¹⁰⁾ to solve the problem of cooling of steel ingots in moulds and of Feild⁽¹¹⁾ who considered the effect of superheat and mould wall thickness on the solidification rate and depth of solidification in terms of time. Lightfoot^(8c,8d) and Russell⁽¹²⁾ have also applied in greater detail the general theory of unsteady heat flow to the solidification of metals and have arrived at more precise results. However, all mathematical methods, including those of C. Schwartz⁽¹³⁾ and Severn^(14,cf.also 15) are open to criticism because of the number of assumptions that have to be made and also because they cannot be applied successfully to alloys that solidify over a freezing range. Nevertheless, one conclusion from the mathematical treatment has proved particularly helpful in interpreting the experimental data, viz. that the velocity of solidification is a parabolic function which can be expressed in the following form:- $d = k\sqrt{t}$

where d depth of solidified metal
 k solidification constant
 t time after pouring.

(b) Direct Quantitative Examination.

In excellent surveys by both West⁽¹⁶⁾ and Ruddle⁽⁹⁾, the work of the previous investigators in this field has been reviewed and compared with the mathematical conclusion stated above.

Nelson⁽¹⁷⁾, Chipman and Fondersmith⁽¹⁸⁾ and also Briggs and Gezelius⁽¹⁹⁾ have found by "pour-out" tests, that

the rate of skin formation agrees well with the theoretical curve at the lower end only, i.e. in the early stages of solidification. Briggs and Gezelius have suggested that the empirical formula: $d = k_1 t^{0.4} + k_2 t$ is a better approximation, the constants depending on the conditions. (As Ruddle has stated, the theoretical curve has been obtained by assuming semi-infinite conditions - a condition that is not satisfied practically.) Briggs and Gezelius have also examined the effect of pouring temperature on the rate of skin formation, the effect of the size of a casting on the rate of skin formation and have measured the approximate temperature conditions, both in the metal and the mould. They found that the sand mould/metal interface remained practically constant at about 1280°C during the solidification process.

Ruddle⁽⁹⁾, like West⁽¹⁶⁾, considered the experimental and theoretical work of Chvorinov⁽²⁰⁾ to be of great importance. Chvorinov studied the rate of heat extraction by sand and other moulds, and found, for example, like Briggs and Gezelius, that the interface temperature remained almost constant throughout the solidification period. (A temperature of 1458°C was given for a 0.24% carbon steel). By calculation he showed that:-

(a) The solidification constant, k , is:-

$$k = 1.158 \frac{e k_s}{L P'}$$

(where $\bar{\theta}$ equals average solidification temperature of steel.
 L equals heat evolved on solidification
 p equals density of metal
 k equals heat diffusivity of the mould

$= \sqrt{\text{thermal conductivity} \times \text{specific heat} \times \text{specific gravity}}.$

(b) The time "t" for complete solidification of a plate is:- $t = (d/2)^2 / k^2$

(where d equals thickness of the plate).

If $d/2$ equals R (where R equals $\frac{\text{volume of plate}}{\text{surface area of plate}}$)

then the equation may be written $t = R^2 / k^2$.

Chvorinov has demonstrated by experiment that, although this equation is only theoretically correct for infinite plates, it may be applied for castings of finite size with only small error. For normal steels, k equals 0.053 if R is in metres, and k equals 2.09 if R is in inches.

Chvorinov criticized the value of pour-out tests for measuring the depth of solidified skin, on the grounds that, by capillary action, metal is retained between the crystal axes, and later solidifies, giving a misleading impression.

West⁽¹⁶⁾ reported that Yearley, Schauss and Martin⁽²¹⁾, and Fifield and Schaum⁽²²⁾, had used the pour-out technique in connection with the solidification of cast iron, and that Troy⁽²³⁾, H.A. Schwartz^(24,25,26) and Clark⁽²⁷⁾ had applied it to cast steel.

A further review is given by Ruddle⁽⁹⁾ of published work on the solidification of non-ferrous castings. The results of these investigations were in general agreement with the conclusions arrived at for ferrous materials.

(c) Observations on the Factors influencing Primary Crystallization.

One of the earliest investigations of the effect of rate of cooling on the formation of dendrites and segregation was made by Priestly⁽²⁸⁾. It was found that with a heavy metal chill, and a rapid rate of cooling, a completely columnar, dendritic structure was obtained, whereas with slower rates of cooling a mixed columnar/equiaxial structure was formed. This changed to a wholly equiaxial type with still slower cooling in a sand mould. He observed that the lower the casting temperature the smaller the crystal size and the finer the segregate. This he considered to be due to a faster rate of formation of nuclei at the lower temperature.

In 1924 the Heterogeneity of Steel Ingots Committee⁸ embarked on a programme of investigations to elucidate the influence of different factors on the structure of steel ingots. The committee investigated the heterogeneity existing in ingots of different types and compositions, with the size varying from a few hundred-weights up to 170 tons. Complete records of the examinations

are given in the different reports, together with possible explanations of the mechanism of freezing.

In the first report^(8a), an account is given of the data obtained from representative ingots. The fundamental facts governing the heterogeneity and mechanism of solidification are discussed fully from the following aspects:-

- (1) Differential solidification, or selective freezing.
- (2) Segregate formation.
- (3) Transition between columnar and equiaxial crystallization, based on a change in the rate of heat extraction, (similar basically to Howes' (6) explanation)
- (4) Gravity settling of crystallites.

A theory based on differential solidification was employed to explain the formation of "free" crystals within the liquid interior of an ingot while solidification was still proceeding at the walls. It was explained that if there is any difference in temperature and composition between the liquid interior and the solidifying wall, then under certain thermal conditions, it is possible for purer crystals to form in the liquid interior. Thus, in the early stages of solidification of an ingot, crystallization proceeds inwards from the walls, the crystals being relatively pure and the temperature being at or only slightly below the liquidus temperature. As steel has a freezing range this leads to the liquid in contact with the growing crystals becoming progressively enriched (in impurities) with the

result that the freezing point is depressed. However, if diffusion is slow and solidification rapid, it is possible for the liquid interior to be purer than the liquid in contact with the growing crystals and therefore to have a higher freezing point. Thus, if there is a shallow temperature gradient between the solidification front and the liquid interior, solidification can occur at the front at one temperature and in the interior at another, slightly higher. In this way "free" crystals may form in the interior of the ingot, and being heavier than the liquid they would then tend to settle towards the bottom. This would result in a central zone of crystals purer than the average metal and in settling they would assume the conical formation commonly observed.

In the second report (8b), there was further discussion of certain features influencing heterogeneity, such as oxides and silicates, sulphur and phosphorus, casting speed and temperature. A study of alloy steel ingots and of non-killed (Rimming) steel ingots was also made.

Later reports (8c,d,e,f,g), while referring principally to the determination of physical data, also considered further aspects of the solidification of steel ingots. Of special importance was the discussion by Northcott^(8f) of the constitution of steel and the influence of differential solidification upon the segregation in steel ingots.

Several other important and valuable contributions were made to our knowledge of the mechanism of dendritic growth and crystallization during the period in which the Heterogeneity of Steel Ingots Committee carried out their investigations (1924 - 1939). The most important of these are mentioned below.

Hultgren⁽²⁹⁾, in discussing crystallization and segregation phenomena in 1.10% carbon steel ingots, suggested a comprehensive theory covering the crystallization of ingots cast under such varied conditions as quiet and turbulent solidification, high and low pouring temperatures, pre-heated moulds, "after-pouring", and vibration. Hultgren's explanation was based upon variations in the degree of under-cooling, depending on the solidification conditions. The transition between columnar and equiaxial growth was also studied and an explanation suggested.

Sauveur and Chou⁽³⁰⁾ examined the influence of rate of cooling on the dendritic and micro-structure of a hypoeutectoid steel. Their conclusions were that in steel which had been slowly solidified and cooled the dendrites were substantially larger than those formed under conditions of more rapid cooling. The formation of a Widmanstätten structure was shown to be related to the cooling conditions.

Matuschka⁽³¹⁾ considered the influence of casting

temperature and the under-cooling capacity of steel. By combining the results of "pour-out" tests with measurements of mould temperatures, he constructed a series of curves indicating the volume and thickness of metal solidified against time. He showed that for a plain carbon steel the freezing velocity was initially high, but fell rapidly to a value that remained constant for some time, and finally increased in the later stages of solidification.

Matuschka considers that at some time during the solidification process a dynamic heat balance is established in which the heat dissipated to the outside of the mould is equal to the heat liberated at the surface of the ingot; the moment at which this occurs has an important influence on the form of crystallization. Thus, columnar crystallization is principally a function of the temperature gradient in the liquid steel, and the character of the crystallization a record of the progress^{ss} of heat transfer. While there is a rapid removal of heat from the interior of the ingot columnar growth is favoured, whereas slower removal of heat, under conditions approaching equilibrium, favouring an equiaxial growth. The precise moment at which the transition occurs depends on factors such as the casting temperature and the under-cooling capacity of the steel. With a lowering of the casting temperature, for example, there is less heat to be dissipated and consequently equilibrium is more rapidly attained.

The results of Matuschka are in general agreement with the views of Hultgren, although his approach to the subject was more mathematical.

Heggie⁽³²⁾ investigated the conditions under which columnar and equiaxial crystals were formed, by measuring the temperature conditions existing during the solidification of stearic acid and tin. He agreed with the view of Matuschka concerning the existence of a state of dynamic heat balance during the solidification process, and likewise concluded that the transition from columnar to equiaxial crystallization was determined by a critical cooling rate. He also concluded that:-

- (a) the critical cooling rate was a function of the width of the freezing range;
- (b) in pure metals, where there is no freezing range, under quiescent conditions only columnar crystals could form;
- (c) motion within the liquid had a pronounced effect on crystallization;
- (d) under-cooling did not necessarily produce equiaxial crystallization.

Northcott^(33,34,35), in a series of papers, has investigated phenomena associated with the crystallization of steel ingots. He has summarized these in a paper on "The Influence of Turbulence on the Structure and Properties of Steel Ingots."⁽⁷⁾ In this paper he suggested a comprehensive explanation of the mechanism of solidification of steel ingots, based on a consideration of the concentration

gradients and of the temperature distribution existing within the solidifying steel. The results of Lightfoot⁽⁸⁰⁾ Matuschka⁽³¹⁾ and Chipman and Fonderamith⁽¹⁸⁾ were reviewed in determining the nature of the probable thermal distribution.

Explanations of typical structures obtained under different conditions were given to illustrate the application of the theory. These are discussed later in the thesis in detail on page 151.

In a similar manner, Northcott has explained periodic⁽³³⁾ and banded⁽³⁶⁾ structures found in centrifugal and other types of casting.

Glazunov and Lazarev⁽³⁷⁾, Barnov⁽³⁸⁾, Scheuer⁽³⁹⁾, Petrosyan and Shou-Shakhbudagyan⁽⁴⁰⁾, Prothero⁽⁴¹⁾ and others have examined the effects of such factors as casting temperature and rate of cooling, and have arrived at similar conclusions to those of Northcott.

Hensel⁽⁴²⁾ and Scheil⁽⁴³⁾ have studied the primary crystallization of metals, and have discussed their findings in relation to the laws governing nucleation and growth. Hensel measured the temperature conditions existing in solidifying tin, and showed that any superheat is absorbed immediately by the mould, the temperature falling rapidly to the liquidus value. He could not detect any under-cooling (to within 0.1°C) or any temperature gradient in the liquid metal after the initial chilling of the metal in filling the

mould. He suggested that convection currents removed any temperature differences.

Scheil concluded that Tammann's laws of nucleation and growth were not applicable to metals. He showed that:-

- (a) The greater the under-cooling of a liquid the less the time required for nuclei to form.
- (b) A high degree of superheat retards the growth of nuclei.

He considered that under-cooling was determined by the rate of nucleation, and by the crystallization velocity; consequently for all metals where the crystallizing velocity is large under-cooling can only be very small.

He further observed that when molten metal solidifies by "skin formation", heat abstraction from the liquid interior must traverse the crystallization front, and as a result the temperature of the interior cannot fall below that of the front.

Like other workers, Scheil found by experiment that the macrostructure of a casting was influenced by the mould temperature, the degree of superheat and the initial degree of under-cooling. However, when the molten metal was superheated to a given temperature, held in the ladle for different periods of time and then finally cast at different temperature, the resulting macrostructure was similar in each case. It was thus concluded that there are "potential nuclei" in the molten metal which have an important influence on the crystal structure of a casting.

Gagen-Torn⁽⁴⁴⁾ likewise concluded that Tammann's Laws apply only to complex organic substances with a low rate of crystallization and not to metals, except when their mass is very small. He also suggested that:-

- (i) growth initially starts from pre-existing (potential) nuclei which are generally small bits of the metal itself;
- (ii) the crystal structure is determined by the rate and direction of heat exchange;
- (iii) the rate of crystallization is determined by the rate of heat extraction and not by under-cooling;
- (iv) movement of metal during cooling affects crystallization by reducing temperature differences and distributing crystals throughout the liquid.

These conclusions are in general agreement with the explanation suggested by Northcott.

Veró⁽⁴⁵⁾ has concluded that variation in the primary structure is not due to variation of rate of cooling or under-cooling. It is thought, however, that the variations in rates of cooling considered by Veró were insufficient to affect a change in structure. He agreed that crystallization was affected by nuclei distribution.

The important role of "potential nuclei" is partially corroborated by thermodynamic considerations. These are discussed on page 27.

(d) Determination of Physical Data:

It is essential for any theoretical study of the

process of solidification of metals that there should exist accurate and reliable information concerning the physical properties of the materials concerned. This need, through the works of numerous investigators, has been partially satisfied, but there still exist many instances where accurate data are lacking.

The important work of the various Iron and Steel Institute Committees should be noted, in particular that of the Heterogeneity of Steel Ingots Committee⁽⁸⁾. They have made many valuable determinations including:-

- (a) the liquidus and solidus ranges in commercial steels;
- (b) the solubility of iron and manganese sulphides in steel;
- (c) the thermal properties and conductivities of ingot mould irons;
- (d) the volume changes in steel from the liquid state to room temperature;
- (e) the fluidity of iron-carbon and other iron alloys.

It is clearly impossible to review all the sources of such information, and where physical data is employed in this thesis the particular source is quoted.

(e) Electrical Analogue Experiments:

Both West⁽¹⁶⁾ and Ruddle⁽⁹⁾ have surveyed the salient features of the electrical analogue method, pointing out that although it is a very valuable method of solving difficult problems of variable heat flow, it is necessary to check the results obtained with direct observations of real castings.

The method, developed primarily by Paschkis⁽⁴⁶⁾, consists essentially of comparing the heat flow in a thermal

system with the flow of electricity in an appropriate electrical circuit. The basic equation of heat flow

$$\frac{\partial \Theta}{\partial t} = \alpha \frac{\partial^2 \Theta}{\partial x^2}$$

is compared with the corresponding equation for the

flow of electrical energy $\frac{\partial E}{\partial t} = \frac{1}{RC} \frac{\partial^2 E}{\partial x^2}$.

Thus the heat content, rate of heat flow, temperature, thermal capacity, and thermal resistivity are equivalent to electric charge, current, voltage, electrical capacity and electrical resistivity respectively.

When the results obtained by the electrical analogues method were compared with those of "pour-out" tests under similar conditions fairly close agreement was found to exist^(47,48). Furthermore, such comparisons have revealed several interesting features of the solidification process.

Paschkis⁽⁴⁷⁾ arranged the electrical circuit in such a way that the effect of air-gaps could be studied during the solidification of steel chill and sand-castings. When these results were compared with Clark's "pour-out" data⁽²⁷⁾ it was found that closest agreement was obtained when an "allowance for an air-gap" was made in the case of chill castings, but not in the case of sand-castings. It was also apparent that during the solidification of sand-castings freezing occurred over a remarkably wide range within the casting. Although it had been generally assumed

that "pour-out" tests referred to some intermediate stage between the beginning and end of freezing, Paschkis' results showed that the "pour-out" tests coincided with the beginning of freeze curves.

Confirmation of a wide freezing range was later obtained by Bishop, Brandt and Pellini from direct temperature measurements (see pages 35,153). The interpretation of "pour-out" tests is discussed in detail on Page 169.

In a further comparison, between the results of Hunsicker's⁽⁴⁹⁾ "pour-out" tests on aluminium sand-castings and the results of the electrical analogue method, Paschkis⁽⁴⁸⁾ the best agreement was obtained without a hypothetical air-gap. As is to be expected from observation, this indicates that sand castings solidify without the formation of an air-gap.

Spretnak⁽⁵⁰⁾ investigated the rate of solidification of cast steel ingots by the electrical analogue method and by "pour-out" tests. He found that if the depth of solidified metal was plotted against the square root of the time, two straight lines could be drawn, representing two stages of solidification. He considered that the intersection of the two lines correspond to the transition from columnar to equiaxial crystallization. Although, as West⁽¹⁶⁾ and Ruddle⁽⁹⁾ pointed out, the evidence is not conclusive, it seems reasonable to assume that a deviation from the initial parabolic relationship might correspond to a change in the type of crystallization, as suggested by Spretnak. Columnar

crystallization occurs under conditions of a steep temperature gradient and narrow freezing range solidification front, whereas equiaxial crystallization occurs when the temperature gradient becomes shallower, with a widening of the freezing range solidification front.

In assessing the merits of the "electrical analogue" method it is apparent, from the results already published, that it is capable of providing valuable information relating to the various unsteady heat flow problems associated with metal solidification. However, these results cannot be fully accepted, without qualification. Ruddle⁽⁹⁾ considers that this is principally due to a limited knowledge of the various thermal properties of materials and to an incomplete understanding of the metallurgical implications.

Consequently it has been necessary to check "analogue" results against "real casts" made under comparable conditions. It is still further unfortunate that most of the information derived from "real casts" has been from "pour-out" tests, the exact significance of which is also in doubt.

However, recent applications of direct temperature measurements for the determination of solidification conditions, furnish data of greater accuracy. This will, in turn, considerably increase the value of the "electrical analogue" method since once the true analogue has been firmly established it will be unnecessary to make individual comparisons with "real casts".

(f) Thermodynamical Considerations.

Birchenall⁽⁵¹⁾, Holloman⁽⁵²⁾, and Turnbull⁽⁵³⁾ in discussing the factors associated with solidification of metals have shown, from thermodynamical considerations, the existence of important relationships between the nucleation and growth of metal crystals and their thermal history and environment.

Birchenall⁽⁵¹⁾ discussed the effect of temperature on the atomic order-disorder transformation of a metal system, and showed how an understanding of this type of reaction could help to explain the phenomena of nucleation and growth. He stated that in a perfect crystal it is possible to specify the position of atoms in the lattice by two types of order - (a) a "short range order" in which the arrangement of neighbouring atoms is determined by the effect of short range inter-atomic forces; and (b) a "long range order" by which the arrangement of groups of atoms is determined over very great distances, i.e. the external form of the crystal. If such a crystal is at the absolute zero of temperature, order is perfect, except for a small "zero point" energy of vibration within the system. However, as the temperature is raised the system expands to accommodate the increased vibration of the atoms, until at a certain stage a few atoms at the surface acquire sufficient energy to move to a different site, or leave the crystal completely. Melting of the metal corresponds to a removal of "long range order"

accomplished by the addition of a considerable quantity of heat to the system - latent heat of fusion. This change occurs at a fixed temperature at which both the solid and liquid phases can co-exist. As Birchenall has pointed out, above this temperature there exists a considerable amount of short range order, but not of long range order, whereas below the melting point the long range order is only disturbed by the presence of a few vacant lattice sites. The dependence of the two types of order on temperature is illustrated in the Figure below.



Although the introduction of another type of atom increases the complexity of the system the feature of the order-disorder transformation remains basically the same.

Holloman⁽⁵²⁾ in considering the theories of homogeneous and heterogeneous nucleation, suggested that the nucleation of molten metals usually takes place heterogeneously, either on the surface of impurities, or on the walls of the container. Impurities act as catalysts when the interfacial free energy between the impurity and the new-phase is less than that between the impurity and the

parent phase. He showed that if steps are taken to remove such impurities, then metals under-cool significantly until eventually conditions for homogeneous nucleation are obtained. The thermal history of the system was also considered to have an important influence on the temperature at which nucleation occurs. He believed that this was due to the retention or elimination of "embryos" of the old phase in cavities in the impurities. With sufficient superheating, the "embryos" are destroyed and undercooling results.

Turnbull⁽⁵³⁾, discussed the consequences of nucleation effects on the kinetics of solidification of metals, and reached several important conclusions, including the following:-

- (a) Crystal form is determined by relative rates of growth in different crystallographic directions rather than by conditions of surface energy.
- (b) "Because of their rapid growth rates, metal crystals can grow isothermally when they are undercooled only slightly. Under conditions of metal solidification, the interface between the crystal and melt is hotter than its surroundings (assumed to mean solid surroundings), and the rate of growth is limited by the speed with which heat can diffuse away from the interface, leading to the well-known pattern of dendritic growth".

From such general considerations it can be seen that there is a distinct relationship between the original laws of nucleation and growth suggested by Tammann, the principle of dynamic heat balance as suggested by Matuschke,

Heggie and others, and the basic theoretical thermodynamic concept of matter. This may lead in the future to a comprehensive explanation of the mechanism of solidification.

2. MORE RECENT INVESTIGATIONS PRIMARILY ASSOCIATED WITH THE SOLIDIFICATION OF SAND CASTINGS.

As a result of a preliminary examination of a series of sand castings, Gray⁽⁵⁴⁾ was one of the first observers to recognize the possible significance of various hitherto unconsidered factors controlling the mechanism of solidification. He realized that although an extensive study had been made of the different aspects of the solidification of steel ingots in metal moulds, relatively little was known of the precise mode of solidification of sand castings. Attention had been directed, as has been shown, by Briggs and Gezelius⁽¹⁹⁾, Chvorinov⁽²⁰⁾, Betty⁽⁵⁵⁾, Taylor and Rominski⁽⁵⁶⁾ and others to the effects of the rate of solidification, and of risers and heads in the production of sand castings. Nevertheless, it was felt that other important factors had not been fully examined - such as the effect of convection currents, gravity, size of section and atmospheric pressure. Gray thought that too much emphasis was being laid, at that time, on the types of head that depended for their efficient operation on the control of the direction of solidification by temperature gradients.

By his investigation Gray was able to show that with different mould arrangements, marked differences in the rates of solidification, types of crystallization and segregation could be obtained. These, he suggested, were due to a large extent to the influence of convection currents,

whose existence initially depended on the shape of the casting and the feeding arrangements.

Realizing the importance of these conclusions the British Iron and Steel Research Association Melting and Metallurgical Committee inaugurated an extensive programme of research to investigate systematically the different factors that might influence the solidification of sand castings. Andrew had⁽⁵⁹⁾ previously expressed the view that such aspects as the feeding of castings should be more fully examined.

The aim of the first investigation⁽⁵⁷⁾ carried out by the Committee was to ascertain the influence of casting temperature and composition upon the structure and properties of horizontally cast steel blocks. It was found that a basic structure existed in all the blocks, consisting of columnar crystallization in the upper half, at right angles to the surface of the mould, and a further two zones in the bottom half: this half consisted of either a lower columnar region (adjacent to the bottom face) separated from the upper half by a randomly oriented equiaxial zone, or a nuclear zone separated from the upper half by the equiaxial zone. A low casting temperature favoured nuclear crystallization in the lower half of the casting. The precise location of the boundaries between the different zones varied with the casting temperature. (As the blocks were cast horizontally, it was considered that the effect of

convection currents was negligible and that changes in structure resulted from a variation of casting temperature alone.) Variation in the carbon content of the steel from 0.12 - 0.54% did not appear to affect the structure.

It was suggested that the differences in the structures found in the two halves of the block were due principally to the gravitational settlement of solid crystallites formed in the upper half: it was assumed that the solid metal was denser than the liquid. This was not the first occasion that such a theory had been proposed; for example it had been suggested (Page 14) that such an action might account for a conical zone of pure equiaxial crystals found towards the bottom of vertical ingots. Benedicks⁽⁶⁰⁾, also, had suggested that the "pedestal" of equiaxial crystals found towards the bottom of Kríž's⁽⁶⁰⁾ vertical sand cast ingots had been formed by crystal fragments settling under gravity from the head. Strong evidence in support of the theory was obtained by inverting a partly solid horizontally-cast block;⁽⁵⁸⁾ a complete reversal of the structure resulted, columnar crystals growing from what had originally been the bottom surface and was now the upper, and nuclear crystals forming along the former upper surface which was now the bottom.

In the second investigation of the British Iron and Steel Research Association Melting and Metallurgical Committee,⁽⁵⁸⁾ the influence of temperature gradients, or

rates of solidification, was examined. This was accomplished by placing metal chills along different faces of the block. Convection currents were again minimized by employing the horizontal form, and to ensure similar composition the blocks were cast from one melt in rapid succession. It was found that chills, although causing faster solidification, did not alter the basic crystalline structure of the blocks; broadly speaking the only effect was to move the centre of freezing away from the geometric centre of the casting. (The centre of freezing was considered to be the junction between the upper columnar and central equiaxial crystallizations.)

In a continuation of this investigation, West determined the temperature conditions existing during the solidification of chill and sand castings, by inserting thermocouples into the mould cavity and measuring the temperature with a high-speed recorder. The temperature gradients and rates of solidification were measured for various degrees of superheat, and correlated with the results of "pour-out" tests. It was concluded that the main factors influencing the form of crystallization were:-

- (1) the distribution of the centres of crystallization in relation to the rate of solidification;
- (2) the settling of crystallites under gravity, and the consequent influence of this on the distribution of the centres of crystallization.

The effects of casting temperature, composition

and rate of solidification having been studied, the next logical step was to investigate the effect of different geometrical arrangements of the castings (as this might introduce convection currents of various intensities and in various directions), while maintaining a uniform composition, casting temperature and rate of solidification. The present investigation was largely concerned with this factor.

Closely related to the above investigations of convection currents, and measurements of temperature conditions existing in sand castings, are the recent investigations of Schaumann⁽⁶¹⁾, Ruddle⁽⁶²⁾ and of Brandt, Bishop and Pellini^(63,64).

Schaumann⁽⁶¹⁾, investigating the effect of convection currents, showed that a Grey Iron casting, bottom poured with sufficient superheat, could be inverted as many as three times, producing temperature gradient reversals within the liquid metal, before the liquidus was reached. This indicated movement of hot liquid metal towards the upper portion of the casting, presumably by convection currents, though turbulence caused by inverting the mould may have affected the result. Temperature reversals were detected in cylindrical castings 10" high with a diameter of $1\frac{1}{8}$ ".

It was found that as the diameter of the castings was increased, less superheat was required to ensure a temperature reversal upon inverting the mould.

Ruddle⁽⁶²⁾ determined by direct measurement the temperature distribution in solidifying aluminium and aluminium alloy sand castings. His thermal results indicated the possible occurrence of two main types of solidification:-

- i) super-purity aluminium, and aluminium bronzes solidified by "skin-formation", according to the parabolic law, whereas
- ii) solid solution alloys having a wide freezing range such as aluminium-4% copper and phosphor-bronze, solidified in a "pasty" manner, with solidification at the centre of the casting occurring concurrently with, although lagging slightly behind, that at the mould walls. During the greater part of the latter type of solidification, liquid and solid existed together in all regions of the casting. Although Ruddle did not correlate these results fully with macro-structure, he observed that pure aluminium, which solidified by "skin-formation", had a wholly columnar crystallization structure; whereas aluminium containing small quantities of titanium solidified in a "pasty" manner, like an alloy of wide freezing range, and had a mixed columnar/equiaxial structure. When cast horizontally, there was a columnar growth from the top of the casting and an equiaxial region at the bottom. It was suggested that the equiaxial structure was produced by the growth of crystallites which had settled under gravity towards the bottom of the casting; this would leave the liquid in the upper part of the mould denuded of nuclei, so permitting unrestricted columnar growth there.

The suggested mode of solidification and resulting structure are akin to those described for horizontal steel castings, and it may be inferred that the thermal distributions were similar.

Bishop, Brandt and Pellini⁽⁶³⁾ studied, by direct thermal analysis, the mechanism of solidification of steel ingots and castings under conditions of different wall thickness, superheat and carbon content. Discussing their results, they drew particular attention to the distinction between the beginning and end of freezing, which they considered to travel as waves across the casting during the solidification process. It was found that in the case of chill castings, with a high rate of heat extraction, the two waves travel in rapid succession across the ingot, creating a relatively narrow band of freezing, whereas in the case of sand moulds, and a slower rate of heat extraction, the "start of freeze wave" had traversed the casting before the "end of freeze wave" had begun, with the result that semi-solid conditions existed throughout the whole casting. Variations in wall thickness, superheat and carbon content were shown to modify the rate of progression of the two "freezing waves". They showed that:-

- (1) the progress of final solidification (i.e. the "end of freeze" wave) obeyed the parabolic law approximately during the early stages of solidification;
- (2) superheat delayed initial solidification and increased the solidification time.

These workers have made valuable contributions to

our knowledge of the subject, but they did not consider their findings in detail in relation to the crystal structure, composition gradient, segregation and distribution of nuclei, and did not discuss the important transition between columnar and equiaxial crystallization.

SECTION TWO.

A METALLURGICAL INVESTIGATION INTO THE EFFECT OF
VARIATION IN THE METHOD OF FEEDING ON THE SOLIDI-
FICATION OF A SERIES OF SAND CAST STEEL BLOCKS.

Synopsis.

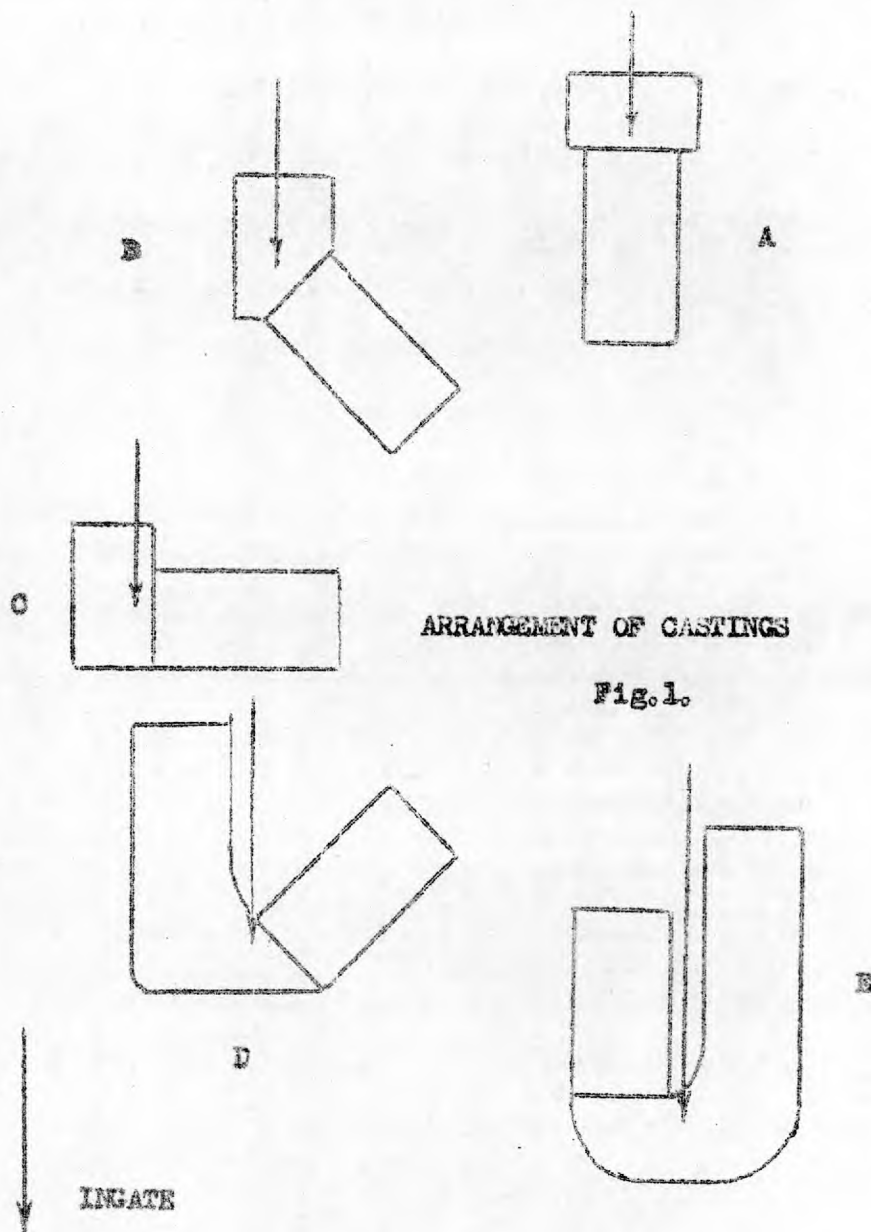
Detailed metallurgical examinations were made of two series of steel castings, each series comprising five blocks, 12" x 6" x 4", which had been cast in different "positions" (i.e. vertically, horizontally etc.) and with different methods of feeding (cf. fig.1).

In the first series, the blocks were cast with the 12" x 6" plane vertical in all cases, while in the second series the 12" x 4" plane was vertical. A normal commercial 0.40% carbon steel was employed, and the castings were made under normal industrial conditions.

Marked differences in crystal structure and segregation were found between the blocks cast in the different position, in each series. The horizontal blocks had the same basic features as those examined in previous investigations^(57,58).

In some castings of the first series solidification from the vertical side walls affected solidification at the centre. This effect was successfully eliminated in the second series.

In the second series the size of the heads of some castings was increased and additions of exothermic



ARRANGEMENT OF CASTINGS

Fig. 1.

compounds were made to the heads, to prevent piping;
this affected the mode of solidification in some cases.

DEFINITION OF TERMS.

A short list of definitions is given in order to avoid any ambiguity concerning the exact meaning of certain words and expressions used to describe different types of crystal. When used in the text, these words will have the meaning given below:-

A dendritic crystal	A crystal which has developed preferentially along certain axes from the nucleus. During solidification regular side branching occurs forming a network, the interstices of which have gradually filled with solidifying metal (upon etching the axes are revealed by differential chemical attack).
An equiaxial crystal	A dendritic crystal of random orientation, having axes of approximately equal length in all directions.
A columnar crystal	A dendritic crystal which has developed preferentially in one direction only along one axis.
A nuclear crystal	A small, cellular, undeveloped crystal which does not reveal a dendritic structure upon etching.

Series I - Experimental Details.

The arrangement of the series of five commercially cast blocks is illustrated in Figure 1 and summarized below:-

A.1. Vertical, top fed.

B.1. Inclined at 45° to vertical, top fed.

C.1. Horizontal, fed at one end.

D.1. Inclined at 45° to vertical, bottom fed.

E.1. Vertical, bottom fed.

The blocks measured 12" x 6" x 4" and were fed by a "whirl-gate" feeder head, as is indicated in the sketches. In this series the blocks were cast with the 12" x 6" plane always vertical.

In order that each casting might be made under the same conditions they were poured from one melt, with the same speed of pouring and in rapid succession. As it appeared possible that convection currents would be most marked in the casting E.1., where feeding was from the bottom, and least in casting A.1., where feeding from the top, it was considered that a comparison of the structure of these two castings would supply information of a very valuable nature: and hence it was essential that they should be cast as far as possible under exactly the same conditions. For this reason E.1. was cast first, then A.1. followed by B.1., C.1., and D.1.

The melt was prepared in a two ton acid electric-arc furnace, following normal commercial practice, at the Grimesthorpe Foundry of the English Steel Corporation. In accordance with the normal practice of the foundry, the tapping and ladle temperatures were determined with an immersion pyrometer prior to teeming, and a quantity of

metal was run from the ladle to warm the nozzle. The metal was deoxidised in the ladle with $2\frac{1}{2}$ lbs per ton of aluminium. For the purposes of the investigation, thermocouples were located in the mould to record the temperature of the incoming metal. Unfortunately, these failed to give satisfactory results, the values recorded being below the liquidus of the steel, but it is thought, in consideration of the large heat content of the ladle and metal, that the temperature drop during the casting of the series of blocks was not appreciable. Probably the drop was only a few degrees, in which case the ladle temperature measured prior to teeming is a close approximation of the true value.

Melting and Casting Particulars.

Tapping Temperature				1610°C
Ladle Temperature (prior to teeming)				1556°C
Metal Composition				
C	0.40%	Mn	0.59%	Si 0.23%
S	0.046%	P	0.046%	Ni 0.19%
		Cr	0.08%	

Liquidus*	1491°C
Superheat (Apparent)	65°C
Nozzle Diameter $1\frac{1}{8}$ " Plumbago stoppers.	
Approximate time for filling moulds	10 - 17 secs.
Deoxidised with $2\frac{1}{2}$ lbs/ton aluminium.	

Examination of Castings.

All castings were examined in the as-cast state. They were sectioned along the central vertical plane of the block, complete with heads. (In this condition the two sets of halves were despatched from the foundry, where they were

* See fig.76, p.118.

cast, to the Royal School of Mines for examination.).

The half containing the section nearest to the centre of the block was turned and machine ground in preparation for sulphur printing and etching. Transverse sections were taken from the other half as required and prepared in a similar manner.

Sulphur prints were taken of the sections: they are shown in Figs. 4, 7, 10, 14 and 18. (An account of the technique finally adopted is given in Appendix I). After sulphur printing, the sections were etched in Oberhoffer's (copper) reagent. When inked the macrostructures were clearly visible, the etching reagent having revealed both the coarse and fine structures present, without excessive pitting. (A description of the technique employed for etching and inking is given in Appendix II.) ~~together with an account of the use of the Loria Etching Reagent~~). In this condition the sections were carefully examined and photographed: the macrostructures of the etched sections are shown in Figs. 3a, 6a, 7, 9a, 12 and 16. Drilled samples for chemical analysis were taken from different crystal zones in order to reveal segregation and perhaps obtain evidence in support of the "falling crystallite" theory. The elements were analysed in the following order: C, S, P, Si, Mn, so that when the sample was small the most important elements from the point of view of segregation were determined. The chemical analyses obtained are shown

in Tables I - IV.

Tensile test pieces were cut from selected positions in the blocks and the following properties determined:-

Ultimate Tensile Strength, Percentage Reduction in Area, Percentage Elongation, Stress for 0.2" Extension.

The results obtained are shown in Tables VI and VII.

Drawings have been constructed illustrating the arrangement and extent of the different types of crystallization present in the castings. The drawings were prepared after a detailed examination of the etched surfaces. The boundaries between the different zones were generally rather indefinite, and where they were unusually vague they are indicated by broken lines in the sketches. As far as possible the types of crystals have been indicated diagrammatically, the general direction of the columnar crystals being indicated by parallel lines. The drawings have been reduced to a third of the actual size.

RESULTS.

Structure of Castings.

Considering the structures revealed in the series of castings, it appears that two extreme modes of crystallization have occurred, typified respectively by the vertical top-fed casting A.1. and the vertical bottom-fed casting E.1. The former casting is composed of the nuclear (cellular) type of crystals (Fig.2), and the latter of the dendritic forms (Fig.16). The other three castings, B.1, C.1, and D.1, have transition structures intermediate between the one type and the other, i.e. between those of castings A.1 and E.1. (Figs.6a, 9a, 12). Consequently, it is most convenient to study the two extreme cases first, and then to examine the changes from A.1. to B.1, B.1 to C.1, C.1 to D.1, and D.1 to E.1.

Casting A.1. (Fig.2). Vertical Top Fed Block.

This block consists of a homogeneous region of nuclear type crystals partially enclosed by a thin envelope of fine columnar chill crystals. The envelope is $\frac{1}{4}$ " thick at the top but decreases and finally disappears toward the bottom. The macrostructure (Figs.3a, 3b) shows that the nuclear crystals are coarser at the top i.e. nearer the head, than those at the bottom of the block.

The sulphur print (Fig.4) shows little evidence of segregation; this is surprising because segregation is

a common feature of vertically cast steel blocks.

Chemical analysis of samples taken from typical positions throughout the block (Table I)* confirmed that there was no noticeable variation in composition.

There is a small area of unsoundness at the top of the block (Fig.3a), caused by the intrusion of the pipe from the head.

* Tables of chemical analysis are to be found after the macrostructures and sulphur prints (Figs.2 - 18).

Casting E.1. (Fig.15) Vertical Bottom Fed Block.

This block consists entirely of coarse dendritic crystals, apart from a small region of nuclear crystals situated in the lower portions of the casting, enveloped by a zone of chill crystals. From the side and top walls the dendritic crystals have a directional tendency at right angles to the wall, which appears to be strongest from the wall furthest away from the feeder head (Figs.16, 17a, 17b). The coarse equiaxial central portion gradually merges with the columnar crystals from the walls. The "pile" of nuclear crystals situated at the base extends from the bottom of the feeder head up into the block (Figs.15, 17c Section V'-V"). At the top of the block a little unsoundness exists.

The sulphur print (Fig.18) shows marked segregation, mainly concentrated in the lower equiaxial portion of the block; it extends down the side of the "pile"

of nuclear crystals nearest the head, but does not penetrate into the nuclear zone. Chemical analysis of samples showed the zone of nuclear crystals to be considerably purer than the crystals in any other region of the block (Table V). This strongly supports the theory of "falling" crystallites, the importance of which is discussed more fully in a later section of the thesis (Page 177).

Casting B.1. (Fig.5). Top Fed Block inclined at 45° to Vertical.

The structure of this block consists mainly of nuclear crystals, like casting A.1., but with some columnar and equiaxial growth.

Below a thin layer of surface chill crystals, columnar crystals extend from the top face towards the centre of the casting. They have developed to a maximum length of about $1\frac{3}{4}$ " near midlength, decreasing near the end wall. They have a pronounced directional tendency, which is approximately 37° from the vertical. Separating the columnar crystals from the nuclear zone at the bottom of the block is a thin region of equiaxial crystals (see longitudinal section Fig.6a). The zone of nuclear crystals occupies the greater portion of the block, and appears to be coarser towards the head than towards the end wall.

The transverse sections BH and BE (Figs.5,6b,6c.)

indicate important features of the structure of the block that are not revealed by the longitudinal section. It is seen that the columnar crystals have developed to their maximum length only along the central plane of the block, whilst the nuclear zone shows a "piled" arrangement resembling that found at the base of casting E.1. A small area between the columnar and nuclear zones consists of mixed transitional nuclear/equiaxial crystals. The equiaxial crystals at the centre are well developed dendritically with secondary and tertiary axes, whilst the crystals near the side walls are more irregular, resembling the nuclear type of crystal. This arrangement is more pronounced towards the head of casting (section BH'-BH'') than at the further end (section BE'-BE'').

The sulphur print (Fig.7) shows that the area of segregate is concentrated in the equiaxial zone at the base of the columnar crystals. The results of chemical analyses (Table II) indicate that the nuclear crystals are purer than either the columnar or equiaxial crystals (compare samples 11, 11a with 9 and 10, and samples 13, 13a with 12), and also that the nuclear crystals nearest the bottom face are purer than those nearer the head (compare 11a and 11 with 13a and 13). This is in accordance with the "falling crystallite" theory to be discussed later.

The area of unsoundness caused by the pipe cavity

is confined to the nuclear area, and not to the area containing the segregate, at the base of the columnar crystals, which is normally thought to be the last portion of the casting to freeze.

Casting C.1. (Fig.8) Horizontal Block, fed at one end.

This block has the same basic structure as was found in similar castings by earlier investigations^(57,58). From the top, side, end, and partially along the bottom wall, there is a layer of chill crystals; this has developed into a regular columnar growth from all but the bottom wall. The longitudinal section (Fig.9a) shows that the vertical columnar crystals are longer than those found in casting B.1. They exhibit a greater uniformity of length and are at right angles to the top wall or cooling surface, except at the corner, where their direction is affected by the conditions of heat loss. They are approximately 2" long.

The nuclear zone is considerably smaller towards the end wall (Figs 8 & 9a), increasing in depth towards the feeder head. The area between the nuclear and columnar zones is occupied by an equiaxial zone. In comparison with casting B.1 (Figs.8, 9b,9c) the transverse sections CE & CH show a more pronounced columnar wall growth, together with a larger area of equiaxial crystals.

The area of segregate (see sulphur print, Fig.10) is confined to the equiaxial zone at the centre of the block.

It is less concentrated than in casting B.1. probably because the equiaxial area is larger.

The chemical analysis results (Table III) again show that the nuclear crystals are purer than either the columnar or equiaxial crystals.

The pipe intrusion, unlike that in casting B.1., is along the centre of the block, and on the same axis as the main segregate. An examination of the pipe area (Fig. 9a) shows that whereas the columnar crystals at the top present a relatively solid front, the equiaxial area is typified by a loose network of randomly oriented crystals showing many cavities.

Casting D.1. (Fig. 11). Bottom Fed Block. Inclined at 45° to the Vertical.

The structure of this block shows a further transition towards the ultimate form of casting E.1.

The arrangement of columnar crystals from the top and end walls is basically the same as that found in casting C.1., although they are shorter ($1\frac{3}{4}$ " from top face), and have less directional tendencies. There is also a layer of chill crystals along all the wall faces. A continuous line may be seen (Figs. 11 and 12) running through the chill crystals at the bottom of the block, parallel to the wall face and turning into the metal adjoining the head.

The nuclear crystals appear to have settled in the lowest portion of the casting, with the equiaxial

crystals occupying the intermediate position between the columnar and nuclear regions.

A section taken in a vertical plane through the junction of the block with the feeder head (Fig. 11, 13c, Section H'-H'') shows that the same type of crystallization are continued into this part of the casting. At the top there is the usual columnar growth. At the bottom there is a layer of columnar and chill crystals, resting on which is a pile of nuclear crystals. The remaining area is occupied by equiaxial crystals which extend down either side of the nuclear "pile". This structure is similar to that shown by section V'-V'' in casting E.1. (Figs. 15, 17c).

The transverse sections (Figs. 11, 13a & 13b) indicate structures which might be expected from an examination of the longitudinal section. In section DE the columnar crystals occupy part of the upper half of the block, the remaining area being occupied by equiaxial crystals only. In section DH a "flattened pile" of nuclear crystals exist along the base of the block.

The area of segregate, as shown by the sulphur print (Fig. 14), is limited to the equiaxial zone in a vague "A" formation and extends high up into the block. No segregate is seen in the nuclear region. The chemical analysis results (Table IV) again support the "falling crystallite" theory by indicating that the nuclear crystals are purer than the other forms of crystallization.

Chemical Analysis.

The results of the chemical analyses, which were carried out at the Bragg Laboratory, Sheffield, are shown in Tables I - V, together with summarized particulars of the positions from which the samples were taken.

The accuracy of the methods of analysis employed were given as:-

C 0.015%; Si 0.01%; Mn 0.01%; P 0.0015%; S 0.002%;

Although the composition differences were mostly within the limits of experimental error, some of them appear to be significant in that definite tendencies are noticeable.

Mechanical Properties.

As the mechanical properties of the different zones of cast steel blocks had been fully investigated in two previous investigations^(57,58), it was considered unnecessary to carry out a similar examination of the present series of castings. However, as the two vertical blocks (A.1. and E.1.) differed so radically from one another as regards crystal structures, it seemed desirable to compare their mechanical properties. Tensile test pieces were therefore machined from the positions indicated in Tables VI and VII, the parallel portions being 1.26" long and 0.357" in diameter.

The results were so variable - perhaps because the material was in the as-cast condition - that no significant differences could be discerned.

Series 2. Experimental Details.

In Series I, where the 12" x 6" plane was vertical, there was some doubt as to whether the side walls had materially affected solidification at the centre of the blocks. For example, the transverse sections of castings B.1, C.1, & D.1 showed that the columnar growth from the upper faces had increased in length, from zero near the side walls to a maximum at the centre. Thus it was not known whether, if the side walls had been further apart, the columnar crystals would have remained at that same length or whether there would have been a further change in structure.

To eliminate this doubt, the second series was cast with the 12" x 4" plane vertical, and the side walls 6" apart. The other details of mould construction and arrangement were similar to those employed for Series I, as summarized in the earlier section.

The melting and casting details were also the same, to allow direct comparison of the structures obtained with those of the earlier series. However, in view of the unsoundness and the existence of pipe cavities in Series I, some of the heads were enlarged and a quantity of exothermic compound was added to each of the feeder heads while still molten. This gave better feeding conditions, but also affected the structures of the castings in some cases.

In this series the temperatures recorded by the

thermocouples placed in the mould cavities were a few degrees above the liquidus of the steel. These temperatures are, however, considered unsatisfactory due to lag and conduction errors, and it is thought that the ladle temperature determined by an immersion pyrometer just prior to teeming, is a more reasonable approximation to the temperature of the metal entering the mould.

Melting and Casting Particulars.

Furnace: Two ton acid electric;
 Tapping Temperature: 1625°C
 Ladle Temperature (prior to teeming): 1570°C
 Metal Composition:
 C 0.33%; Mn 0.72%; Si 0.31%; S 0.041%;
 P 0.038%; Ni 0.18%; Cr 0.10%;

Liquidus* 1496°C
 Superheat (Apparent) 74°C
 Nozzle Diameter 1½"
 Approximate time for filling moulds: 13 - 15 seconds
 Deoxidized with 2½ lbs/ton aluminium.

Examination of Castings.

The castings were sectioned, machined and ground in the same manner as was adopted for the earlier set.

No samples were taken for chemical analysis, neither were tensile test pieces prepared, as this was considered unnecessary in the light of the results of the previous series.

As before, a series of drawings was prepared from the actual etched surfaces, illustrating diagrammatically the different zones of crystallization present in the castings.

*see fig.76, facing p.118.

RESULTS.Structure of Castings.

A preliminary comparison of the structures of castings A2 and E2, in which the intensity of convection currents would be expected to be at a minimum and maximum respectively (Figs 19,31), shows that two radically different structures were again produced. However, in this series the casting fed at the top, A2 (Fig.19), has a structure which is entirely dendritic, whereas in Series I, casting A.1. was wholly nuclear (Fig.2); furthermore, the casting fed at the bottom, E.2., is almost wholly nuclear (Fig.47), whereas in Series I casting E.1. was composed mainly of coarse dendritic crystals (Fig.15). There is, therefore, an unexpected complete reversal of the types of structures. Differences also exist in the other castings of the series, as a more detailed examination will show.

Casting A2 (Fig.19) Vertical Top Fed Block.

This casting consists entirely of dendritic crystals. There are columnar crystals extending to a maximum length of approximately 1" from the side walls. The central portion, between the columnar zones, consists of randomly oriented equiaxial crystals which are coarser towards the top of the block than at the bottom (Fig.20a).

The transverse section (Figs.19,20b), shows that there is a similar columnar growth from the other two walls, with the equiaxial crystals occupying the central area. This section also shows that the columnar crystals are of

uniform length across the width of the block, indicating that the side walls have not affected the crystallization at the centre.

The areas of segregation shown in the sulphur print (Fig. 21) are distributed in the equiaxial zone in the form of a double isosceles triangle ' Δ ', the areas near the apexes appearing more concentrated than those towards the bases. The apparent concentration near the apexes is probably related to the greater depth of growth of the columnar crystals in this part of the casting.

When the two castings A1 and A2 including the block and feeder head are compared, it can be seen that A2 has a considerably larger head; in fact the volume of the head of A2 is more than twice that of A1. This has prevented a pipe extending into the block, but has also, apparently, affected the mode of crystallization.

Casting E.2. (Fig. 31.) Vertical Bottom fed Block.

The structure of this casting consists almost entirely of nuclear crystals except for a roughly rectangular area of coarse dendritic crystals situated towards the top of the block (Figs. 31, 32a).

Transverse sections show that the nuclear zone extends across the block at position E2H (Figs. 31, 32c), and totally envelopes the dendritic zone at position E2E (Figs. 31, 32b). The nuclear crystals do not appear in the

usual globular form but jointed in small, fine, stringers which are inclined towards the vertical, the arrangement being somewhat similar to a " $\wedge$ " segregate. Two coarser nuclear areas extend down towards the lower portions of the block from the corners of the dendritic zone, parallel to the side walls, enclosing an area of fine nuclear crystals between them.

The dendritic zone (measuring 2" x 3" x 4") is enclosed completely by approximately $\frac{3}{4}$ " to 1" of nuclear crystals. This zone is further sub-divided into a region of columnar crystals, which are at right angles to the side and top walls, and a small central area of randomly oriented equiaxial crystals.

There are pronounced segregate areas concentrated in stringers which assume the position shown in the sulphur print (Fig. 51). One of the stringers towards the top of the block appears to 'invade' the dendritic area, the axes of the dendrites appearing to pass completely through it (see also Fig. 48).

Further comments on the structure of this casting, which is abnormal in many respects, will be made later (page 186).

Casting B2. (Fig. 22). Top-fed Block inclined at 45° to the vertical.

In casting B2, columnar crystals have grown from all walls. They extend to a depth of $1\frac{1}{2}$ " from the

top wall near the head, and to a depth of 2" near the end wall. Along the entire length of the bottom wall the columnar crystals have developed to a depth of about $\frac{1}{2}$ ". The central area between the upper and lower columnar zones consists of an equiaxial/nuclear zone; this is mainly equiaxial near the end wall, and composed of coarse nuclear crystals towards the head.

A close examination of the columnar growth from the upper face shows that discontinuities or disturbances have occurred at two positions (Fig. 23a); one position is half way down the block and the other towards the feeder head. In each case the width of the disturbed area is greater towards the centre of the block than at the wall. The general appearance suggests that the columnar front has been disturbed and a new form of crystallization has developed.

The transverse sections B2E and B2H (Figs. 23b, 23c) show that the side walls have not affected the crystallization at the centre, the columnar crystals from the top face having reached a uniform length in the central portion. Columnar growth is seen to extend from the side walls, modifying the shape of the central equiaxial-nuclear zone.

The sulphur print (Fig. 24) shows "V"-type segregation; the main concentration is at the base of the

columnar crystals, but the segregate also extends into the equiaxial/nuclear area.

Casting C2. (Fig.25). Horizontal Block, fed at one end.

The structure of casting C2 has the same basic features as Casting C1 (Fig.8), and also resembles that of casting B2.

Along the top wall there exists a regular zone of columnar crystals extending to a maximum depth of rather more than 2". There is also a columnar formation from the end wall. The nuclear zone along the bottom face thickens slightly towards the end wall. Between the columnar and nuclear crystals in an area of randomly oriented equiaxial crystals.

As in casting B2, the transverse sections (Figs.26b,26c), show that the side walls have not affected crystallization at the centre of the block. In contact with the side walls chill crystals have developed into columnar crystals, the remaining space being occupied by equiaxial and nuclear zones.

The sulphur print (Fig.27) shows that the segregate is mostly confined to the junction of the columnar and equiaxial zones.

There is no pipe cavity in the block, owing to the enlarged feeder head.

Casting D2 (Fig.28), Bottom fed Block inclined at 45° to the Vertical.

There is a marked similarity between the structure of this block and the corresponding casting of the earlier series, Casting D1.

In contact with the top, bottom and side walls, there is a layer of chill crystals, developing from the top wall into regular columnar crystals which extend for nearly 2" towards the centre of the block. There is also columnar growth from the end wall. In contact with the chill crystals along the bottom face, there is a zone of nuclear crystals, which is thicker near the feeder head than near the end wall. Their arrangement suggests that they have "settled" into this position. The area between the columnar and nuclear crystals is occupied by randomly oriented equiaxial crystals. (See also Fig. 29a).

As in castings B2 and C2, the transverse sections (Figs. 28, 29b, 29c) show that solidification from the side walls has not affected freezing at the centre of the block. In the upper half of the block, columnar crystals have developed from the side walls, whilst the lower portion is occupied by nuclear/equiaxial formations.

The areas of segregate (Sulphur print Fig. 30), like those of casting C2, are concentrated at the junction of the columnar and equiaxial zones, extending in some places into the equiaxial region.

Preliminary Comments on the Results of the Examination of the Two Series of Industrial Castings.

A comparison of the structures of castings Al and El shows that although the liquid metal was cast under exactly the same conditions, the arrangement of the feeder head has had a profound effect on the resulting mode of crystallization. In the top-fed casting Al, a completely nuclear crystalline structure was obtained, whereas casting El, which was bottom fed, had a coarse dendritic columnar/equiaxial structure. According to Gray⁽⁵⁴⁾, convection currents should be at a minimum in the top-fed casting and at a maximum with bottom feeding. If so, it would seem that convection currents may have a considerable effect on the mode of crystallization, perhaps by maintaining a continuous supply of heat from the head.

Examination of the intermediate castings Bl, Cl and Dl shows that there was a gradual transition from the nuclear structure of casting Al to the dendritic structure of El, and this might be ascribed, at least in part, to the effect of increasing convection currents.

It must be pointed out, however, that the evidence provided by the examination of this series of castings does not prove that the structural changes brought about by altering the arrangement of the block and position of the feeder head was actually caused by convection currents. There is an alternative explanation, which has much in its favour,

and which will be discussed later, in the section starting on page 134, after additional evidence has been put forward.

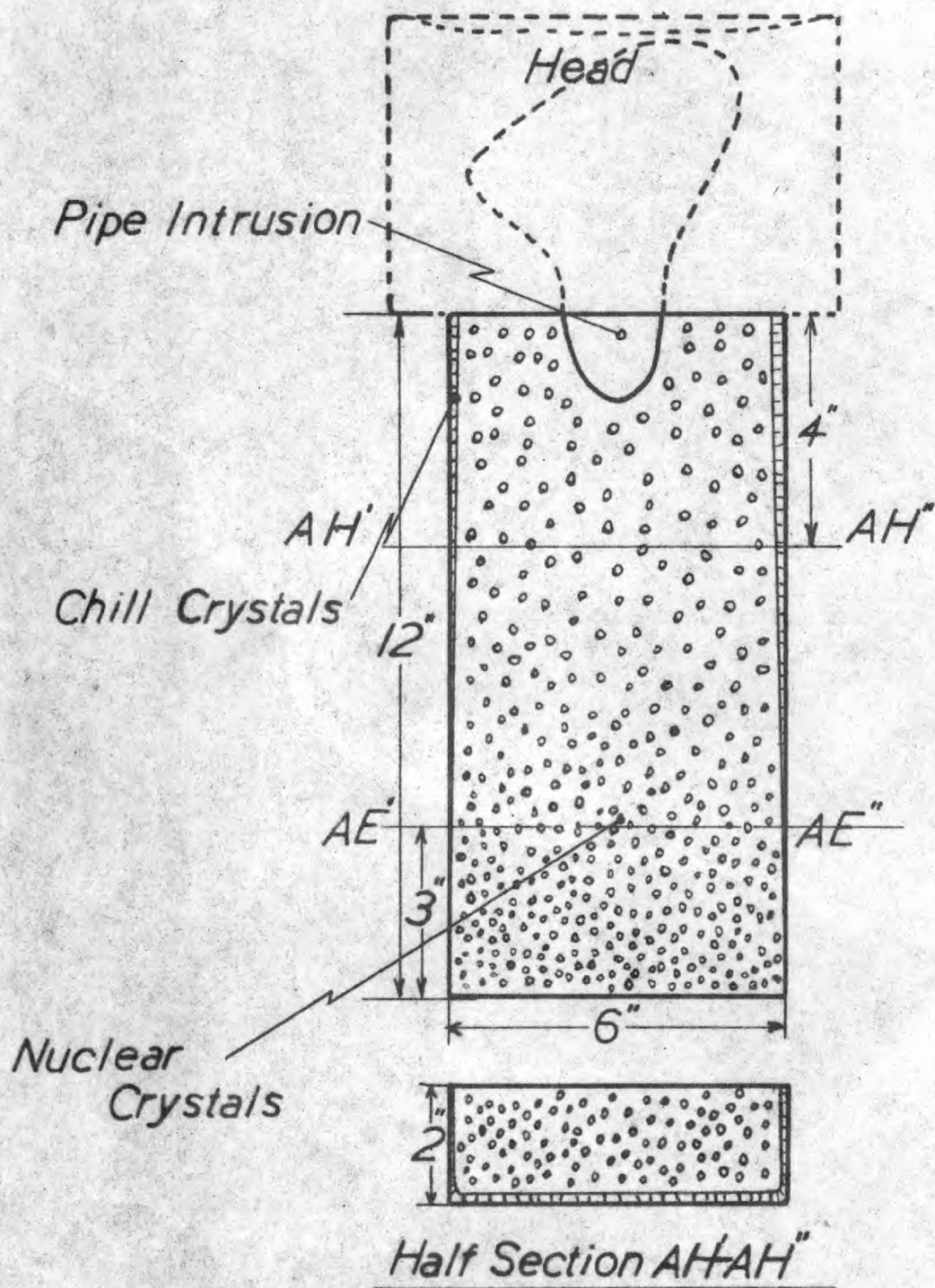
In all castings, it appears that the nuclear crystals have settled towards the bottom of the blocks; this is discussed in detail on Page 179.

Examination of the sulphur prints shows that the areas of segregate are situated in the central equiaxial zone near the head, and do not extend into the columnar or nuclear zones. These segregate areas were presumably the last portions of the blocks to freeze. In the nuclear casting A1, the structure appears very homogeneous and does not reveal any areas of segregate. In casting E1, as has been noted on page 44, the areas of segregate extend from the central equiaxial area into the head down one side of the "nuclear pile".

The three intermediate castings in the second series (B2, C2 and D2) have structures broadly resembling those of the corresponding members of the first series, but the two vertically cast blocks (A2 and E2) have radically different structures from those of the corresponding blocks (A1 and E1) of the first series. The difference between the two top-fed blocks (A1 and A2) may reasonably be ascribed to the fact that A2 had a much larger head than A1 and that additions of an exothermic compound were made to A2 and not to A1. There does not, however, seem to be

any reasonable explanation of the abnormal structure of casting E2, except the assumption that something went seriously wrong with the experiment. If this is accepted, there must be some doubt as to the significance of all differences between corresponding members of the two series. Because of this doubt, it was considered advisable to make a similar series of castings in the laboratory. The size of the blocks had to be reduced considerably, and it was impossible to avoid some variation in chemical composition from casting to casting, because the only suitable melting furnace available had a maximum capacity of about 20 lbs, but it was felt that ability to control the casting conditions accurately and to vary them as required outweighed the disadvantages.

Further discussion of the structures of the industrial castings is therefore postponed to section 4 of the thesis (beginning on page 134) where they can be considered in the light of the laboratory experiments.

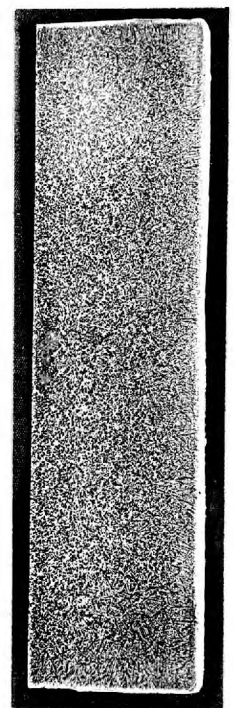


STRUCTURE OF INGOT A (12"×6"×4")

FIG. 2



(a) Longitudinal Section



(b) Transverse
Section AH'-AH''

Fig 3. Macrostructure of Casting A1.

HEAD

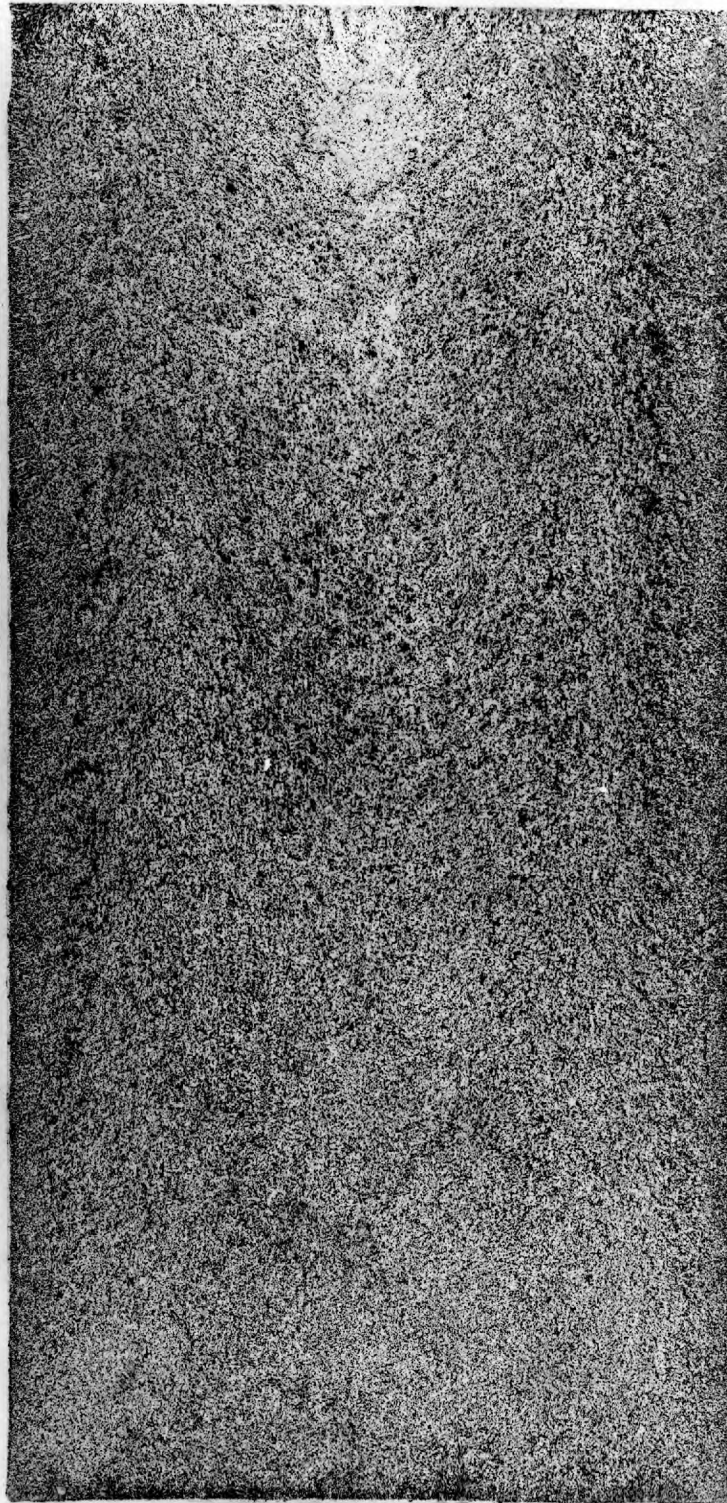


Fig. 4. Sulphur Print. Casting Al.

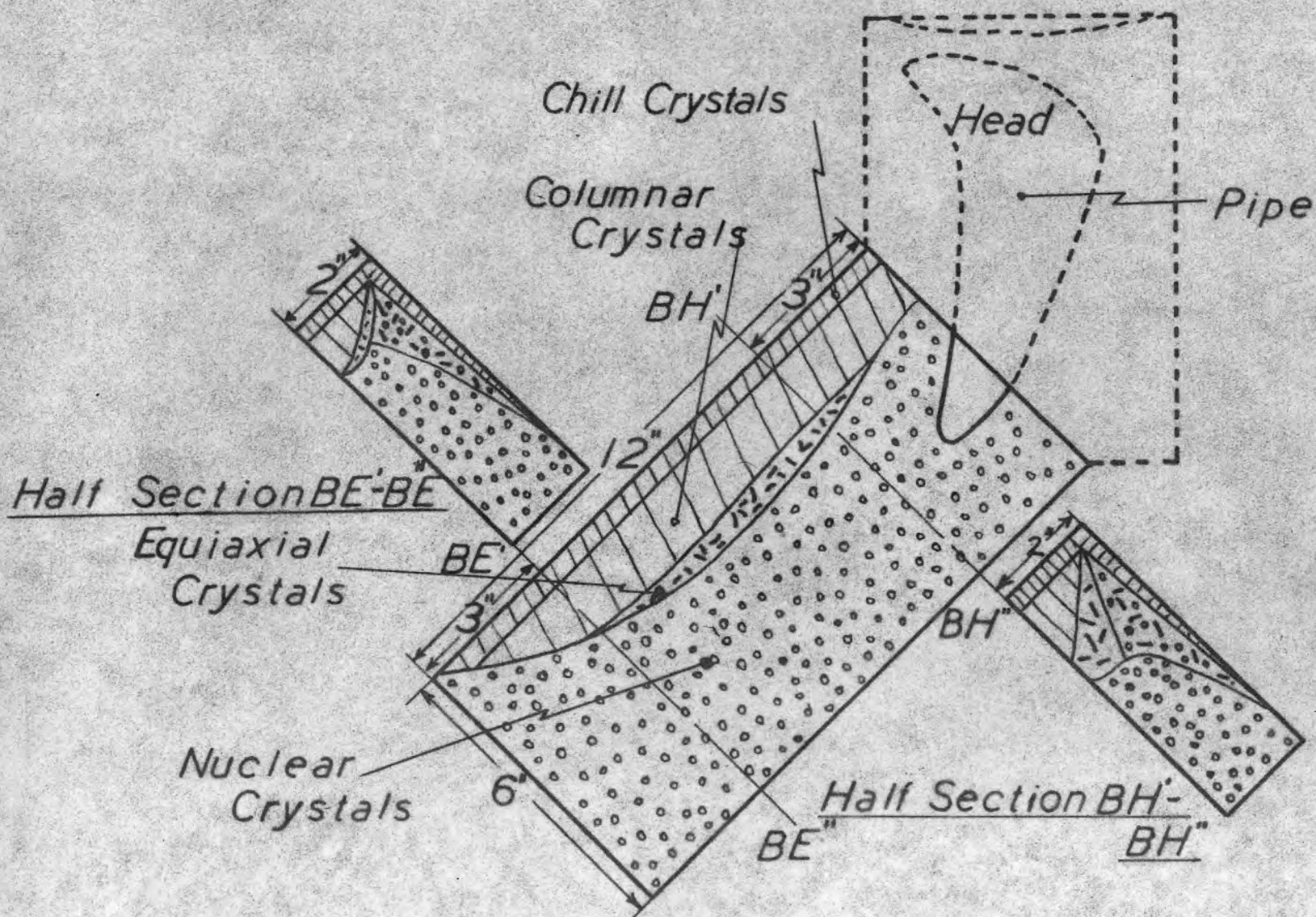
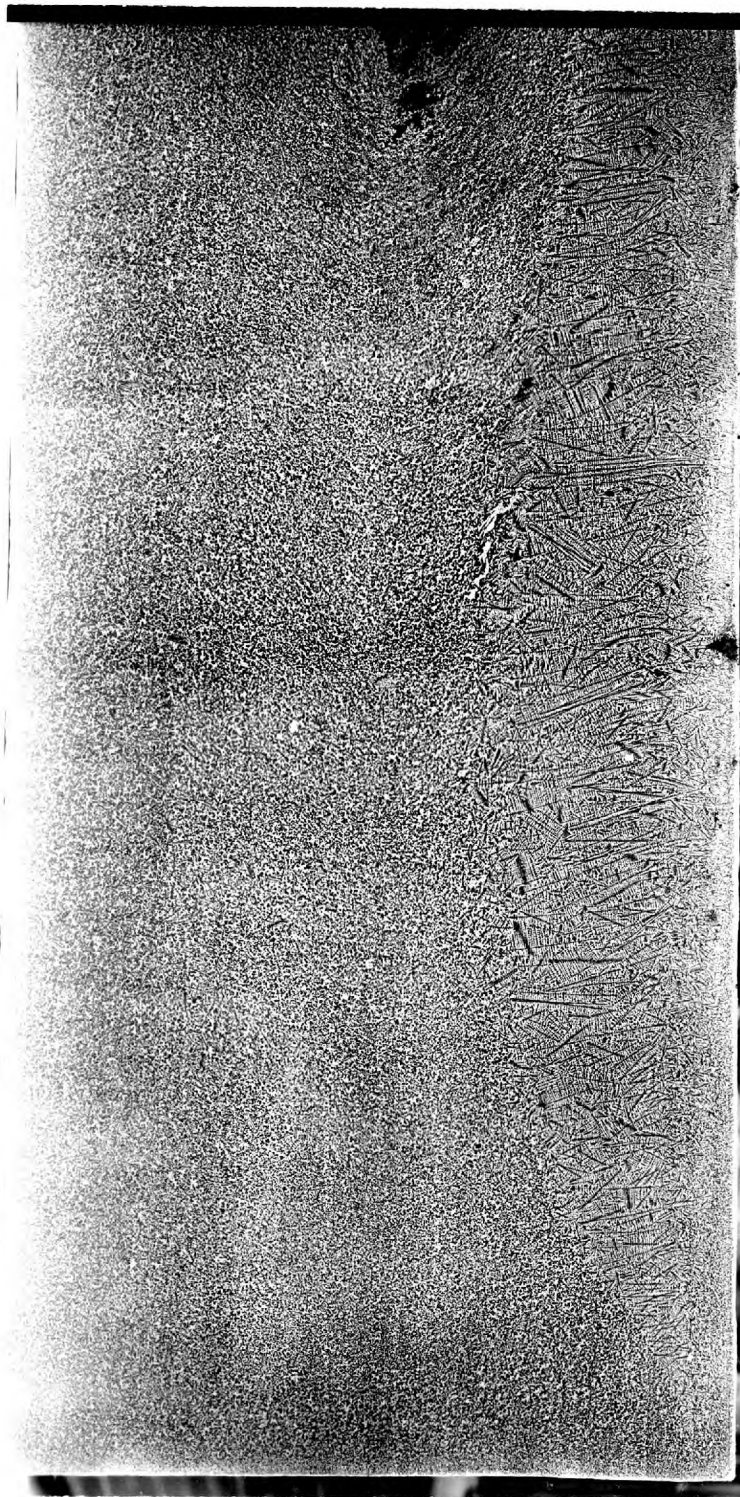


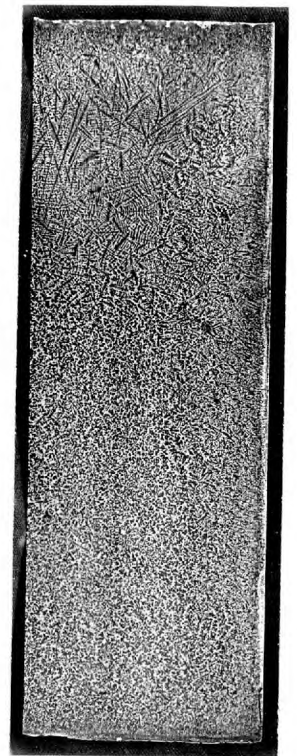
FIG. 5

STRUCTURE OF INGOT B (12x6x4)

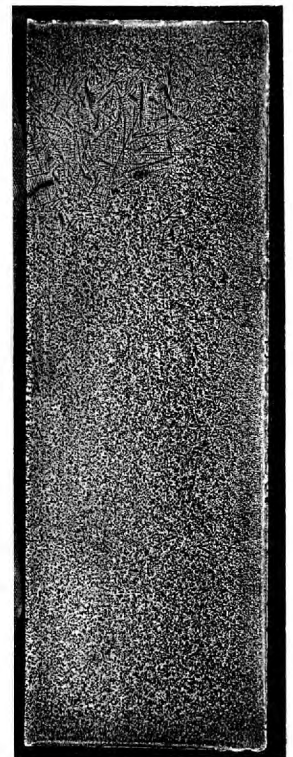
HEAD



(a) Longitudinal Section



(b) Transverse
Section BH'-BH''



(c) Transverse
Section BE'-BE''

Fig. 6. Macrostructure of Casting B1.

HEAD

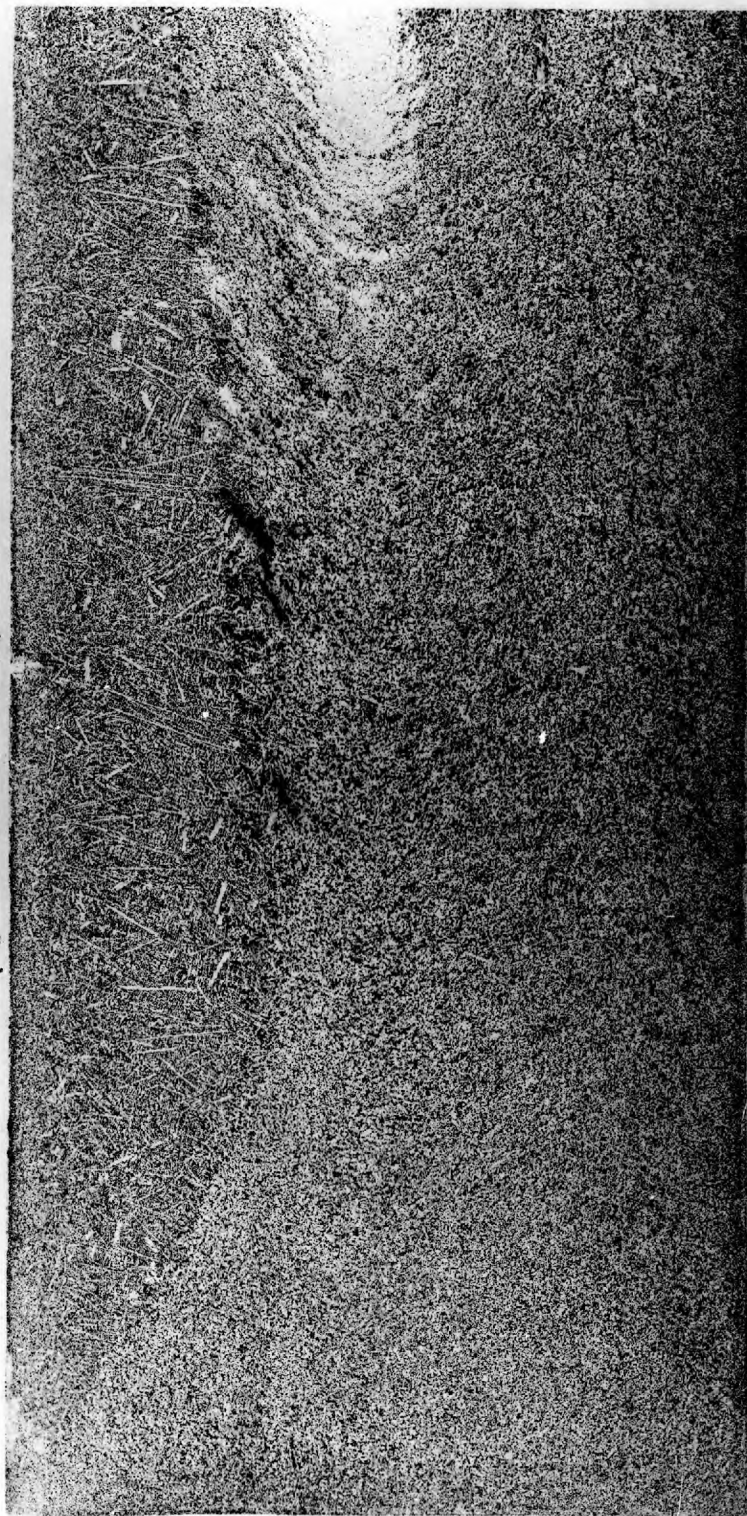


Fig. 7. Sulphur Print. Casting Bl.

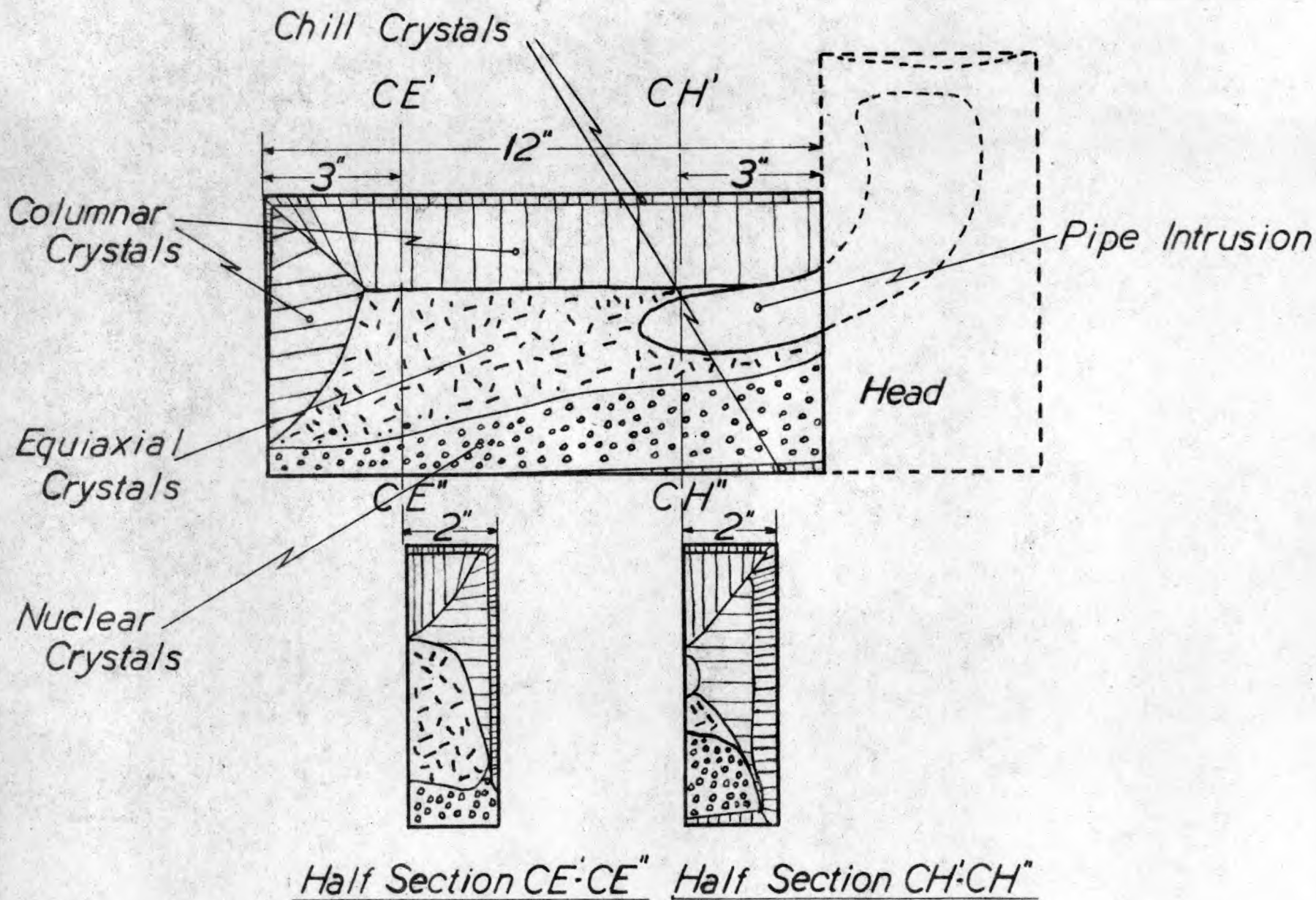
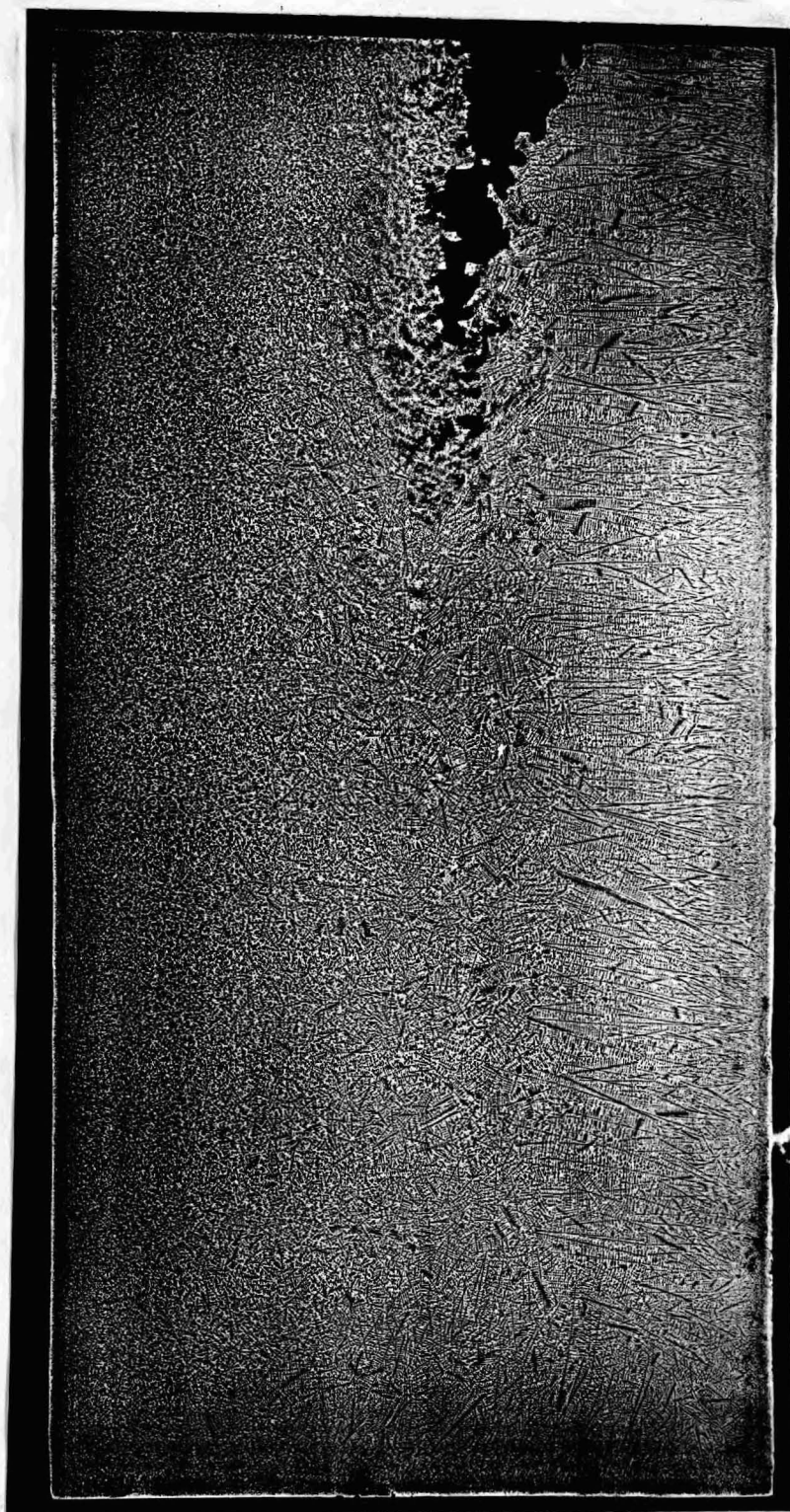


FIG. 8 STRUCTURE OF INGOT C (12" x 6" x 4")

HEAD



(a) Longitudinal Section



(b) Transverse
Section $CH'-CH''$



(c) Transverse
Section $CH'-CH''$

Fig. 9. Macrostructure of Casting Cl.

HEAD



Fig. 10. Sulphur Print. Casting Cl.

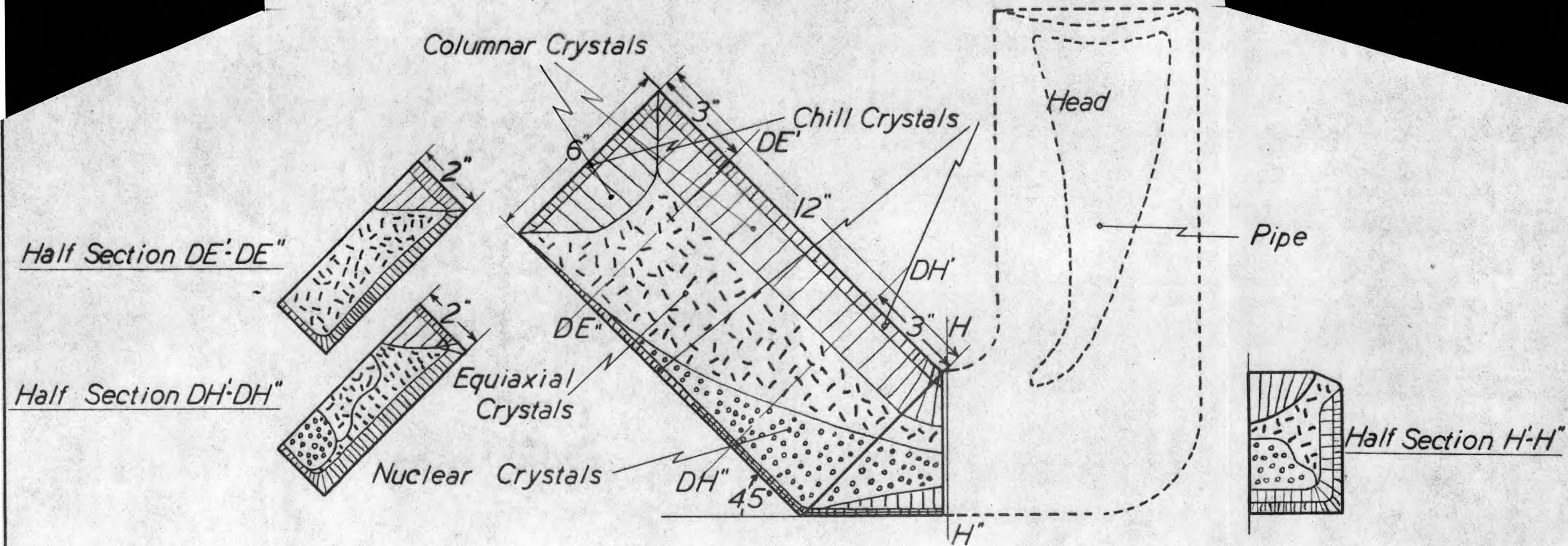
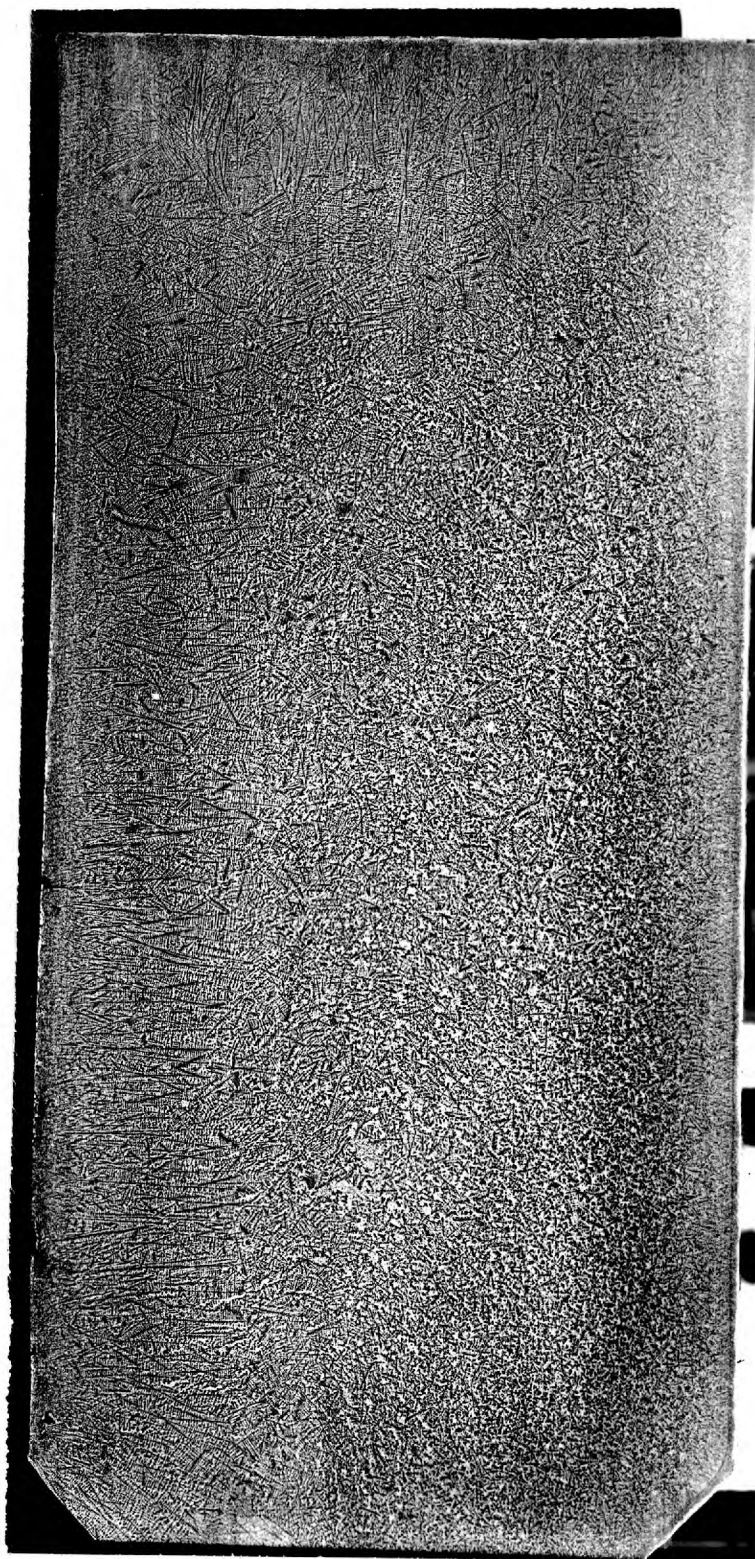


FIG. II STRUCTURE OF INGOT D (12" x 6" x 4")



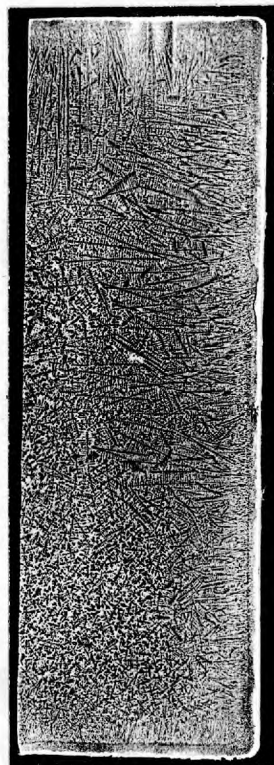
HEAD

Longitudinal Section

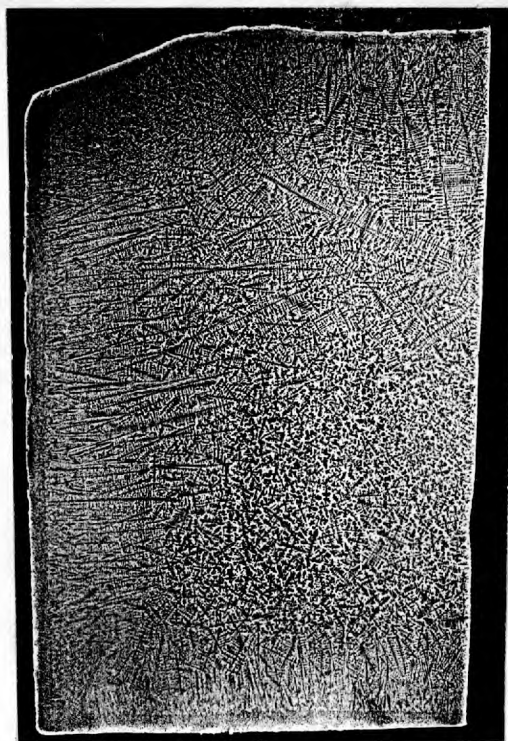
Fig. 12. Macrostructure of Casting D1.



(a) Transverse
Section DE'-DE''



(b) Transverse
Section BH'-BH''



(c) Section H'-H''

Fig. 13. Macrostructure of Casting D1.



HEAD

Fig. 14. Sulphur Print. Casting D1.

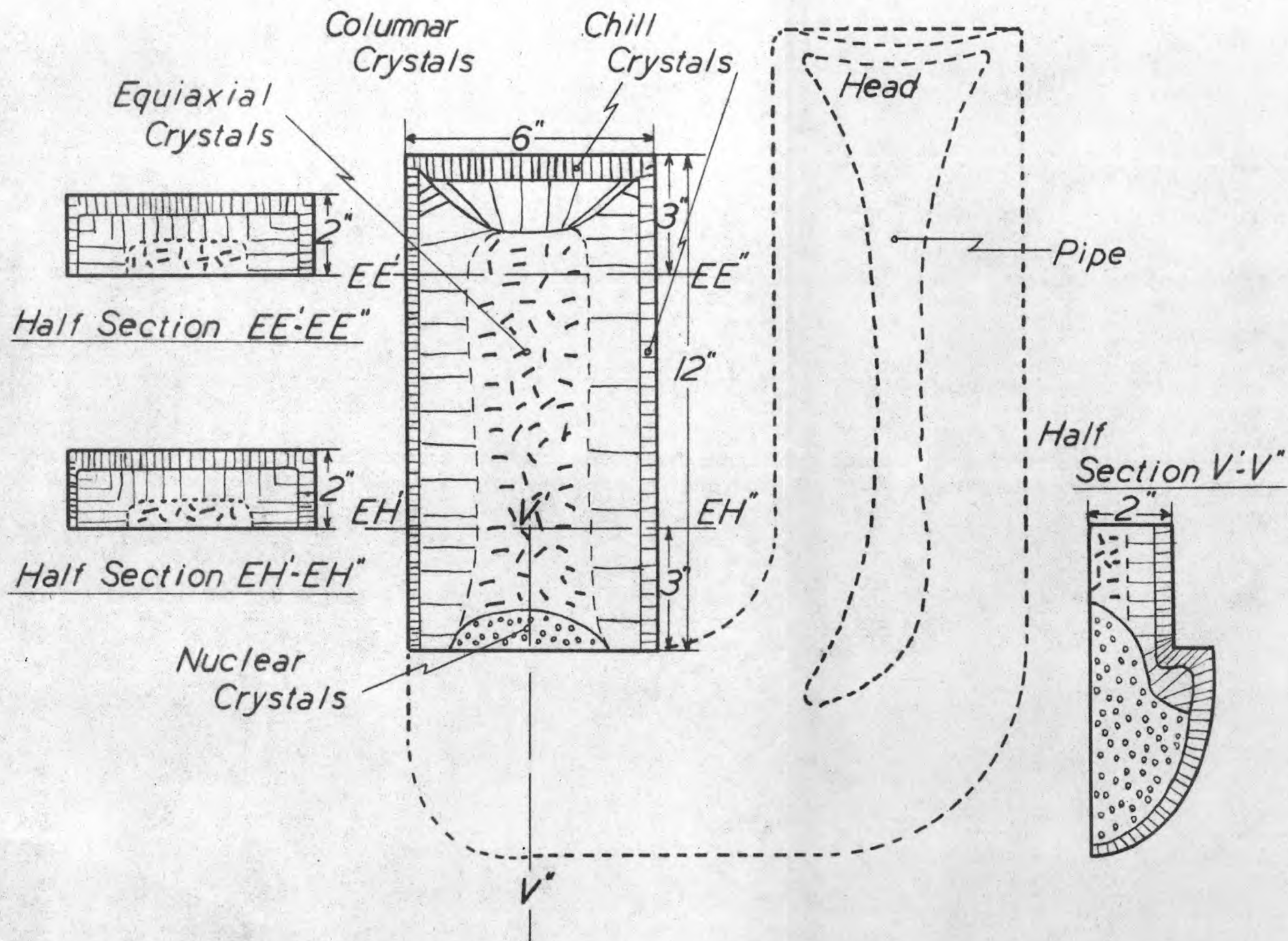
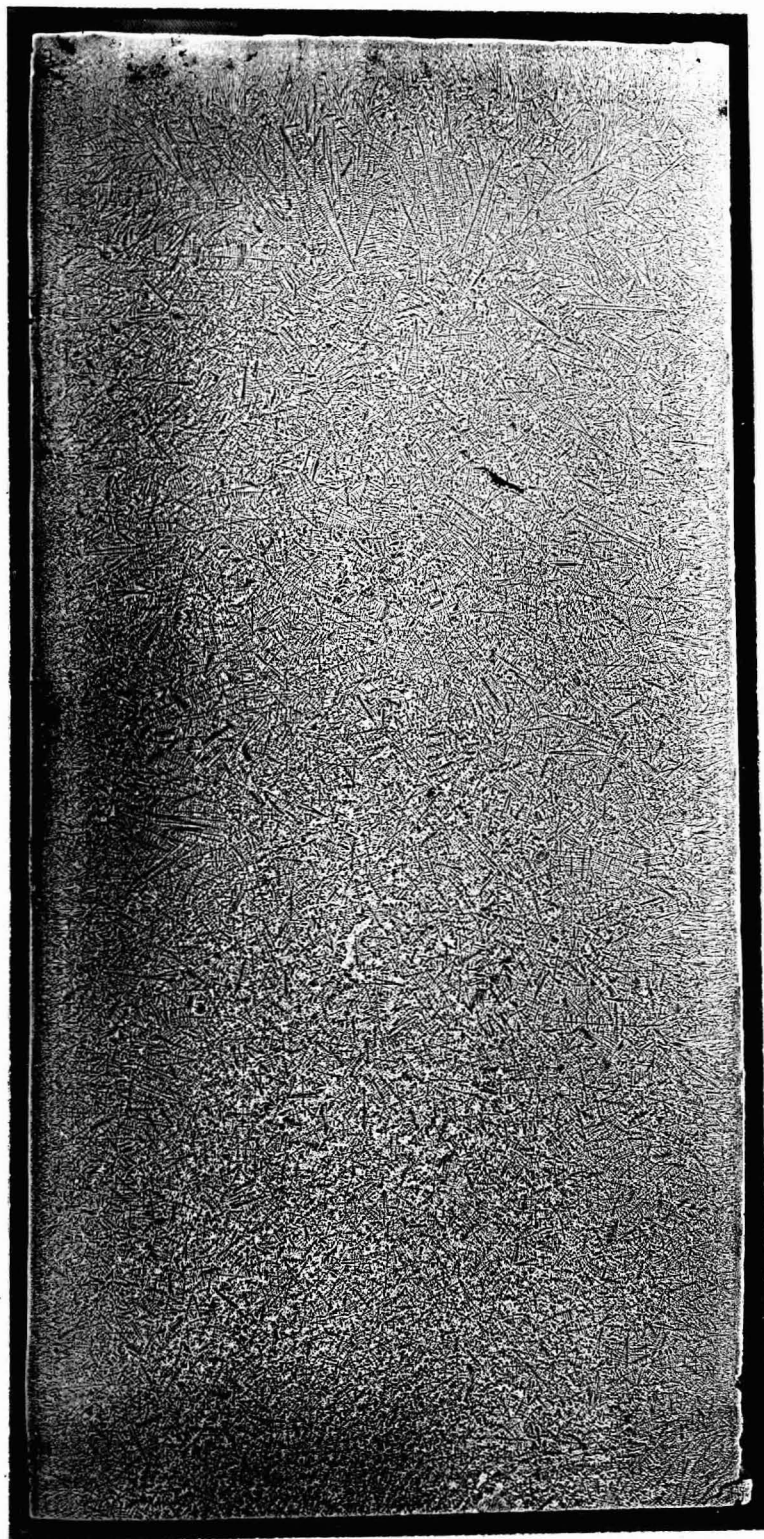


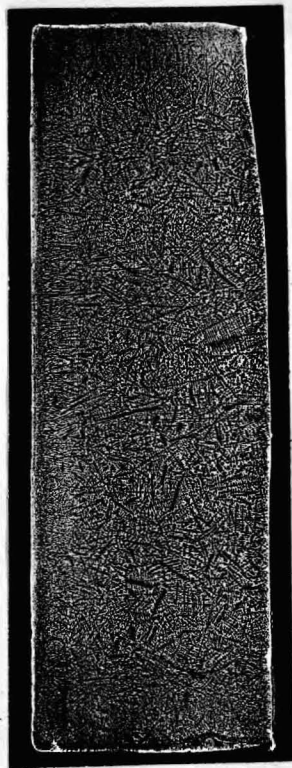
FIG. 15 STRUCTURE OF INGOT E (12" x 6" x 4")



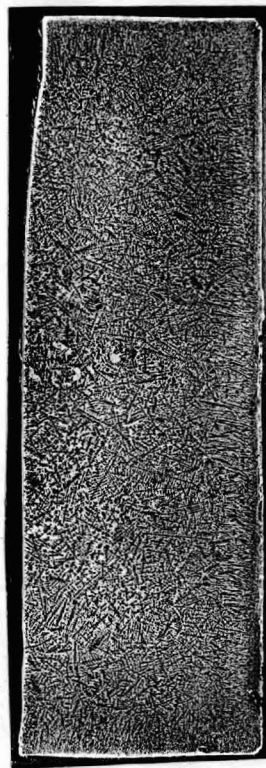
HEAD

Longitudinal Section

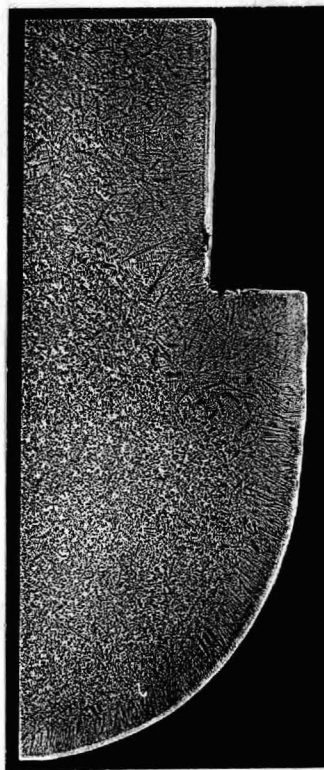
Fig. 16. Macrostructure of Casting El.



(a) Transverse
Section EE'-EE''



(b) Transverse
Section EH'-EH''



(c) Section V'-V''

Fig. 17. Macrostructure of Casting El.



HEAD

Fig. 18. Sulphur Print. Casting El.

RESULTS OF CHEMICAL ANALYSES.

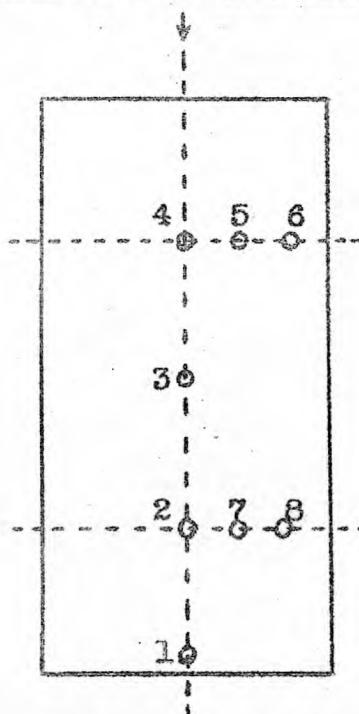


TABLE I. Chemical Analysis Casting Al.

1. Centre line, $\frac{1}{8}$ " from bottom.	C 0.38: S 0.048: P 0.048.
2. Centre line, 3" from bottom.	C 0.39: S 0.044: P 0.044.
3. Centre line, $7\frac{1}{2}$ " from bottom.	C 0.39: S 0.046: P 0.048.
4. Centre line, 3" from top.	C 0.36: S 0.033: P 0.038.
5. Line 3" from top, 1" from centre.	C 0.36: S 0.046: P 0.044: Si 0.24.
6. Line 3" from top, 2" from centre.	C 0.38: S 0.047: P 0.046: Si 0.24.
7. Line 3" from bottom, 1" from centre.	C 0.40: S 0.045: P 0.047.
8. Line 3" from bottom, 2" from centre.	C 0.38: S 0.048: P 0.047.

ALL NUCLEAR CRYSTALS.

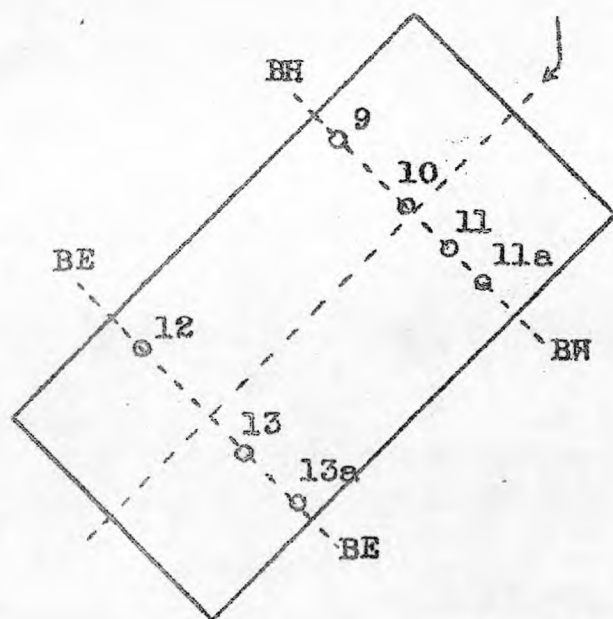


TABLE II. Chemical Analysis Casting Bl.

9. Section BH 3" from head, $\frac{3}{4}$ " from top face.
Columnar crystals.
C 0.42: S - P 0.052.
10. Section BH 3" from head, $2\frac{3}{4}$ " from top face.
Equiaxial crystals.
C 0.44: S 0.043: P 0.052.
11. Section BH 3" from head, 4" from top face.
Nuclear crystals.
C 0.42: S 0.044: P 0.046
Si 0.022.
- 11a. Section BH 3" from head, 1" from bottom face.
Nuclear crystals.
C 0.40: S 0.038: P 0.045:
Si 0.23: Mn 0.51.
12. Section BE 3" from end, 1" from top face.
Columnar crystals.
C 0.42: S 0.052: P 0.051.
13. Section BE 3" from end, 2" from bottom face.
Nuclear crystals.
C 0.39: S 0.044: P 0.046:
Si 0.24.
- 13A. Section BE 3" from end, $\frac{7}{8}$ " from bottom face.
C 0.37: S 0.039: P 0.045:
Si 0.24.

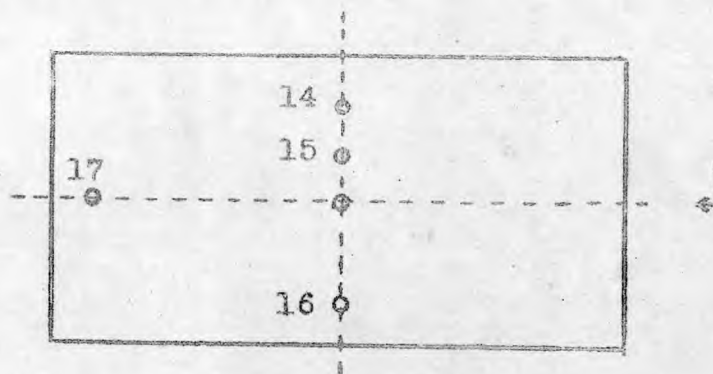


TABLE III. Chemical Analysis Casting Cl.

- | | | |
|-----|---|--|
| 14. | Section 6" from end, $1\frac{1}{2}$ " from top.
Columnar crystals. | C 0.43: S 0.045: P 0.049. |
| 15. | Section 6" from end, $3\frac{1}{2}$ " from top.
Equiaxial crystals. | C 0.41: S 0.045: P 0.046. |
| 16. | Section 6" from end, $\frac{3}{4}$ " from bottom.
Nuclear crystals. | C 0.38: S 0.044: P 0.045:
Mn 0.54: Si 0.24. |
| 17. | Section in Columnar crystals at end wall halfway down face, $\frac{7}{8}$ " from end. | C 0.40: S 0.044: P 0.045:
Si 0.25. |

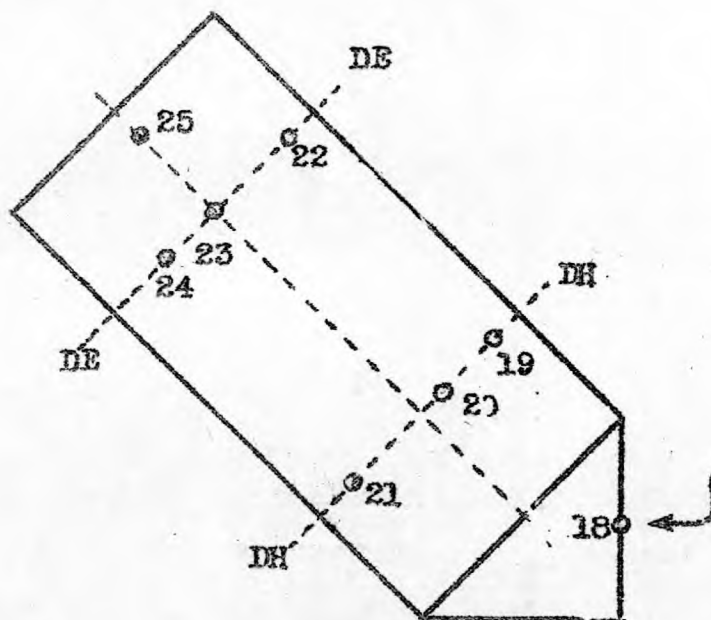


TABLE IV. Chemical Analysis: Casting D1.

- | | |
|---|--------------------------------------|
| 18. Vertical section at base: $2\frac{3}{4}$ " from top. Equiaxial crystals. | C 0.41: S 0.047: P 0.050 |
| 19. Section DH 3" from feeder end: $\frac{3}{4}$ " from top. Columnar crystals. | C 0.41: S 0.045: P 0.045
Mn 0.54. |
| 20. Section DH 3" from feeder end: $2\frac{3}{4}$ " from top. Equiaxial crystals. | C 0.42: S 0.046: P 0.046
Mn 0.55. |
| 21. Section DH 3" from feeder end: 1" from bottom. Nuclear crystals. | C 0.40: S 0.044: P 0.048 |
| 22. Section DE 3" from end: $\frac{3}{4}$ " from top. Columnar crystals. | C 0.40: S 0.044: P 0.044 |
| 23. Section DE 3" from end: 3" from top. Equiaxial crystals. | C 0.43: S 0.044: P 0.047 |
| 24. Section DE 3" from end: $1\frac{9}{16}$ " from bottom. Equiaxial crystals. | C 0.41: S 0.042: P 0.045 |
| 25. On centre line, $\frac{3}{4}$ " from end. Columnar crystals. | C 0.39: S 0.041: P 0.044 |

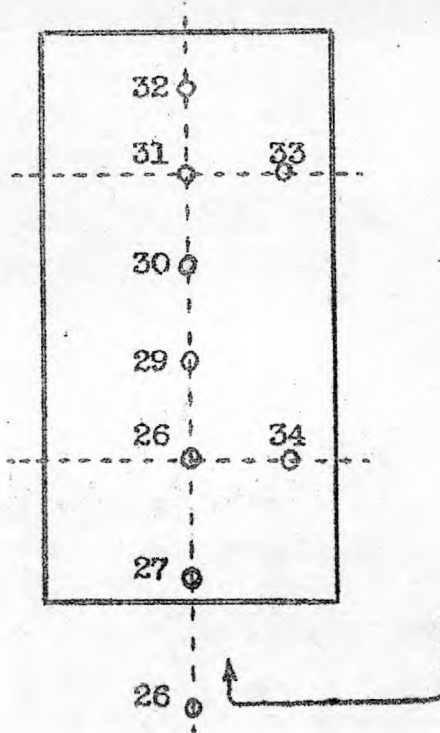
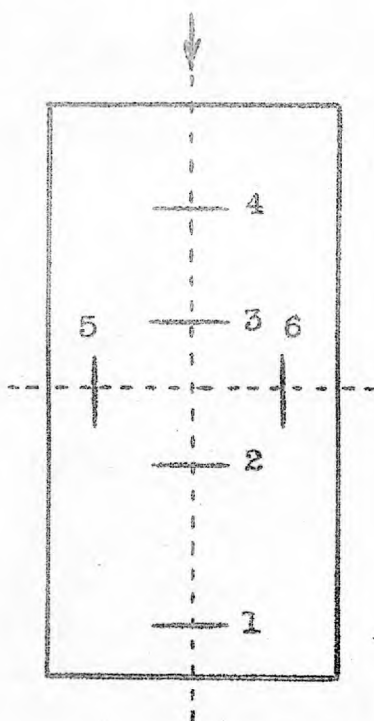


TABLE V. Chemical Analysis : Casting El.

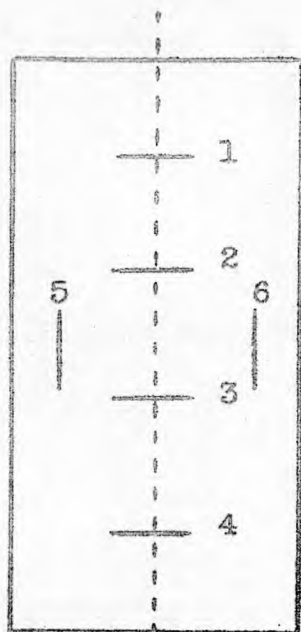
26. Centre line: $2\frac{1}{4}$ " below casting. Nuclear crystals.	C 0.038: S 0.042: P 0.044: Mn 0.55.
27. Centre line, $\frac{1}{8}$ " from bottom. Nuclear crystals.	C 0.36: S 0.039: P 0.042.
28. Centre line, Section EA 3" from bottom. Equiaxial crystals.	C 0.43: S 0.044: P 0.049.
29. Centre line, $4\frac{1}{8}$ " from bottom. Equiaxial crystals.	C 0.46: S 0.049: P 0.049: Si 0.24: Mn 0.00.
30. Centre line, 7" from bottom. Equiaxial crystals.	C 0.46: S 0.049: P 0.049.
31. Centre line, Section EE 9" from bottom. Equiaxial crystals.	C 0.42: S 0.043: P 0.047.
32. Centre line, $1\frac{1}{4}$ " from top. Columnar crystals.	C 0.39: S 0.044: P 0.048.
33. Section EE 9" from bottom, 1" from wall near to head. Columnar crystals.	C 0.40: S 0.044: P 0.048.
34. Section EH 3" from bottom: 1" from wall near to head. Columnar crystals.	C 0.42: S 0.046: P 0.051: Si 0.25.



1. - 1" from bottom of casting.
2. - $4\frac{1}{2}$ " -do-
3. - $7\frac{1}{8}$ " -do-
4. - $2\frac{1}{4}$ " from top of casting.
5. - 6" -do- - 1" from left wall.
6. - 6" -do- - 1" from right wall.

TABLE VI. Mechanical Properties : Casting Al.

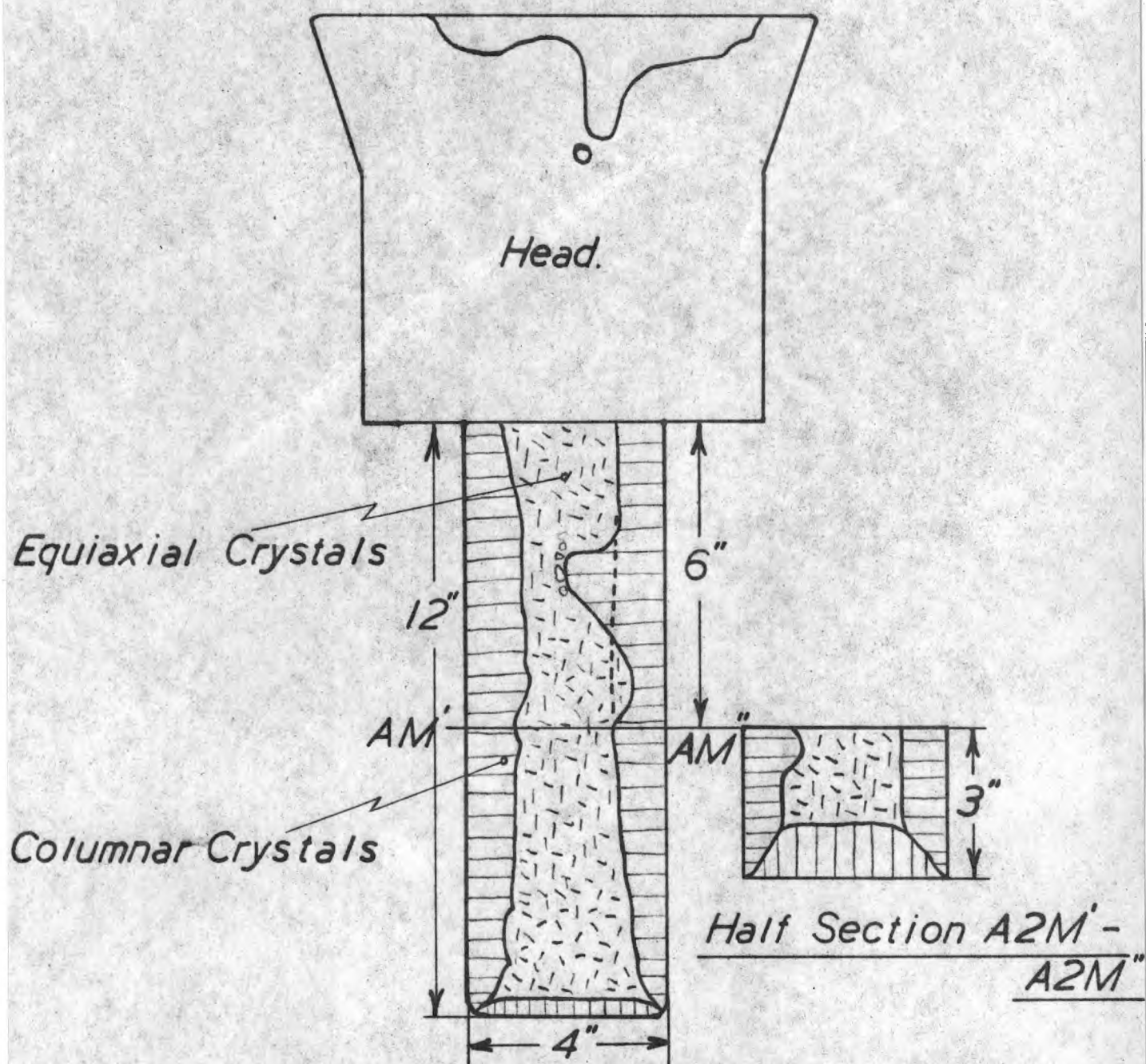
Specimen	Ultimate Tensile Strength. Tons per sq.in.	Percentage Reduction in area.	Elongation %	Stress at 0.2" extension Tons per sq.in.
Al (1)	32.0	9.83	7.65	20.9
" (2)	29.6	6.63	4.65	21.1
" (3)	31.9	8.40	6.80	18.12
" (4)	23.3	6.60	4.82	17.0
" (5)	33.0	7.14	5.67	21.9
" (6)	33.2	7.20	7.14	19.4



1. - 2" from top of casting.
2. - 4 $\frac{1}{2}$ " -do-
3. - 7" -do-
4. - 2" from bottom of casting.
5. - 6" from top of casting - 1" from left side.
6. - 6" from top of casting - 1" from right side.

TABLE VII. Mechanical Properties : Casting El.

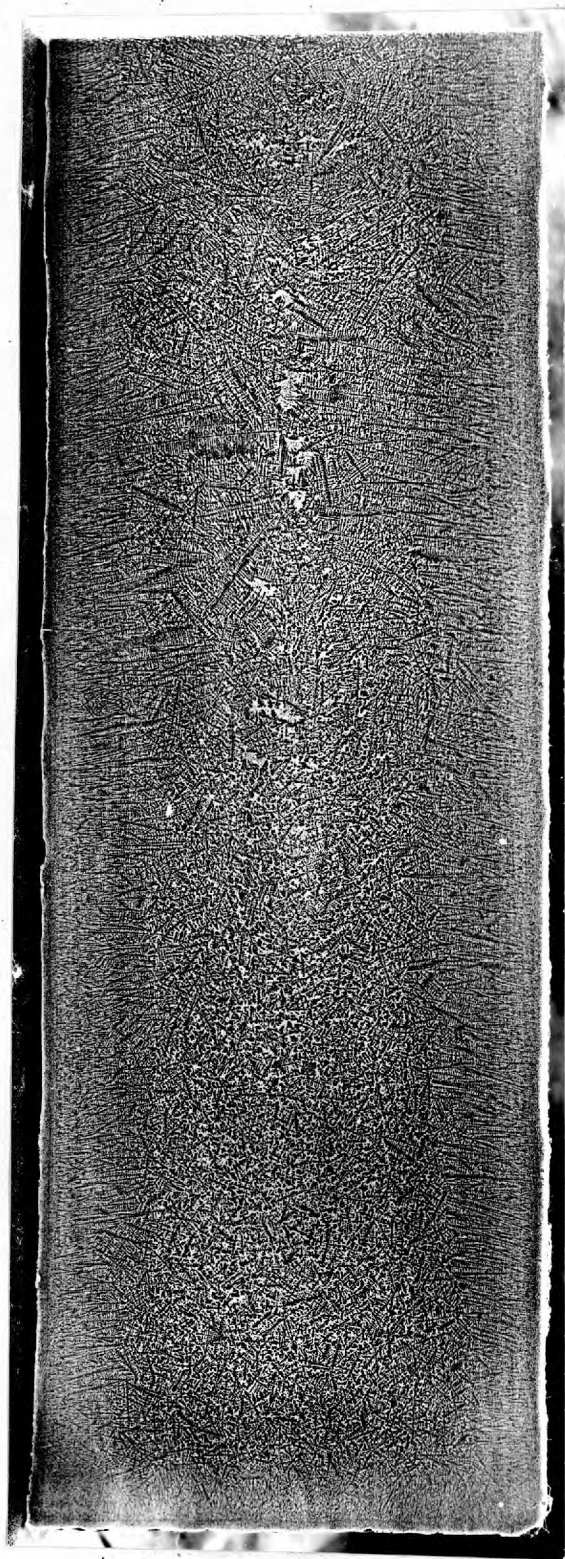
Specimen	Ultimate Tensile Strength. Tons per sq.in.	Percentage Reduction in area.	Elongation % %	Stress at 0.2" extension Tons per sq.in.
El (1)	32.48	7.65	7.37	17.2
" (2)	30.3	5.00	4.65	20.4
" (3)	28.9	3.90	4.48	18.9
" (4)	-	-	-	-
" (5)	33.5	6.63	7.79	18.76
" (6)	34.4	6.63	8.59	19.30



STRUCTURE OF CASTING A2.

FIG.19

HEAD



(a) Longitudinal Section



(b) Transverse Section A2M'-A2M''

Fig. 20. Macrostructure of Casting A2.

HEAD

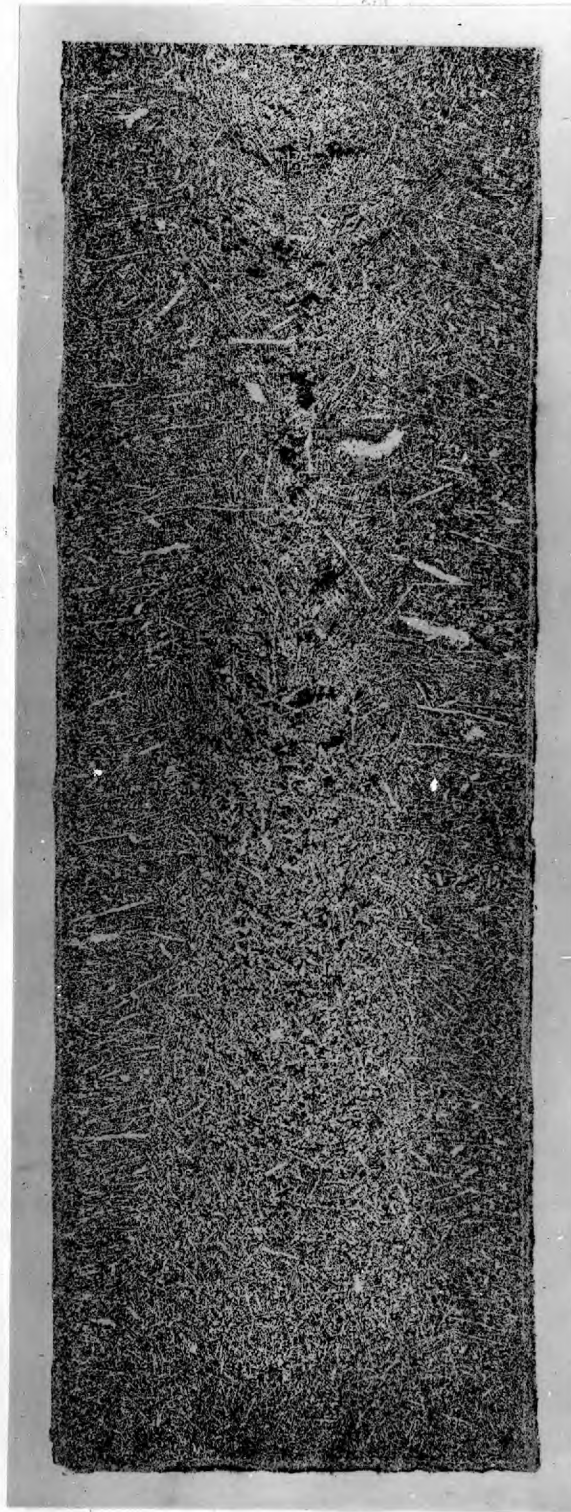


Fig. 21. Sulphur Print. Casting A2

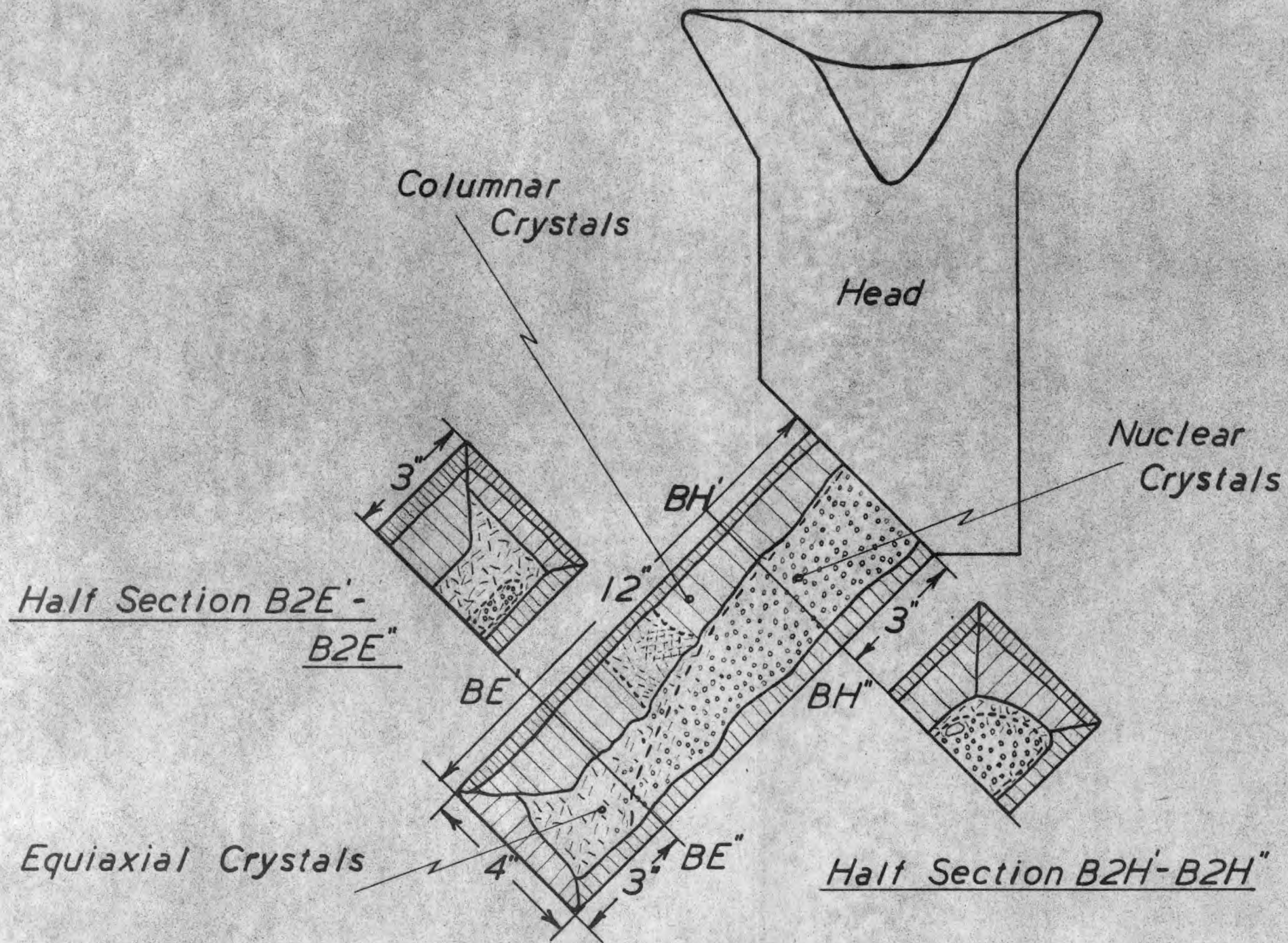
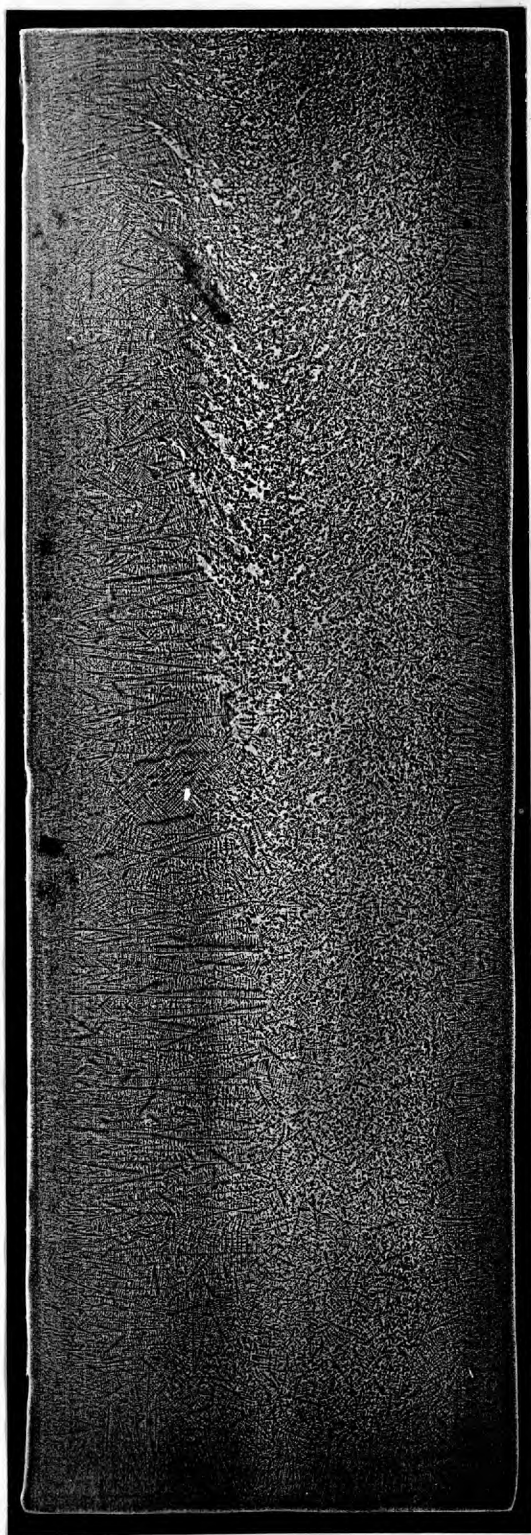
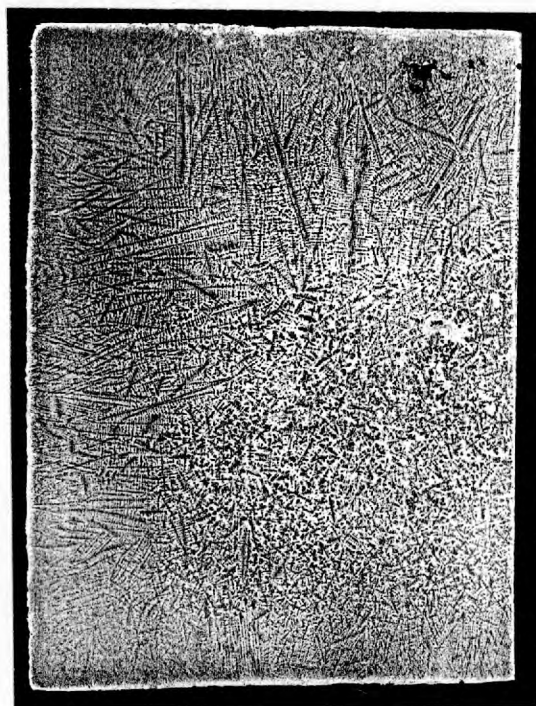


FIG. 22 STRUCTURE OF CASTING B2.

HEAD



(a) Longitudinal Section



(b) Transverse Section
B2H'-B2H''



(c) Transverse Section
B2E'-B2E''

Fig. 23. Macrostructure of Castings B2.

HEAD

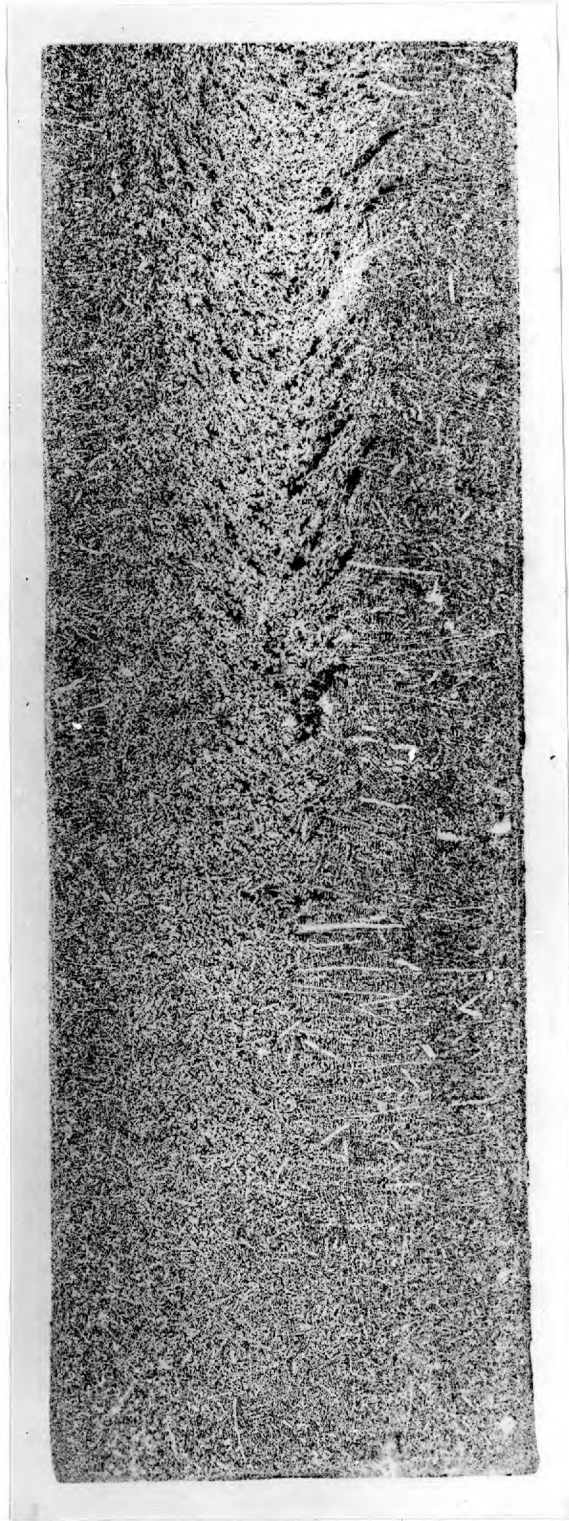


Fig. 24. Sulphur Print. Casting B2.

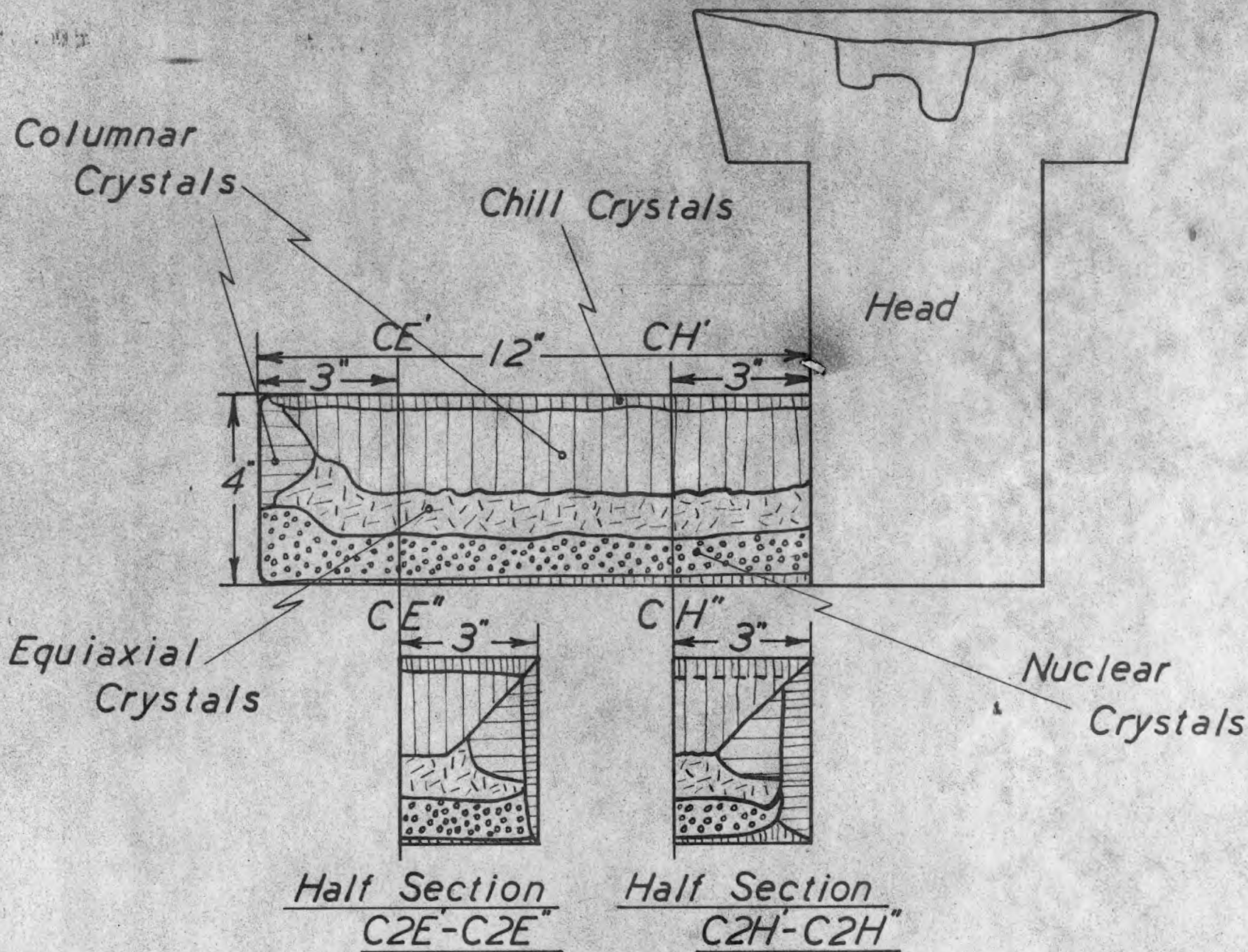
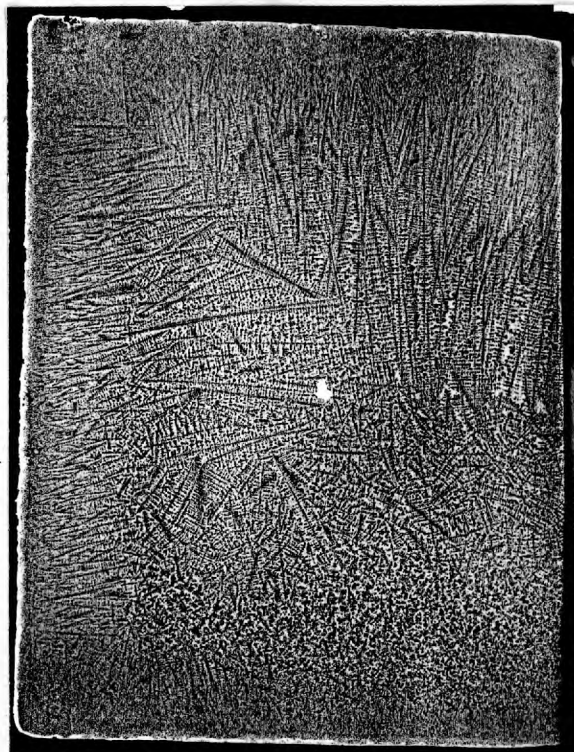


FIG.25

STRUCTURE OF CASTING C2.

HEAD



(b) Transverse Section C2H'-C2H''



(a) Longitudinal Section (c) Transverse Section C2E'-C2E''

Fig. 26. Macrostructure of Casting C2.

HEAD

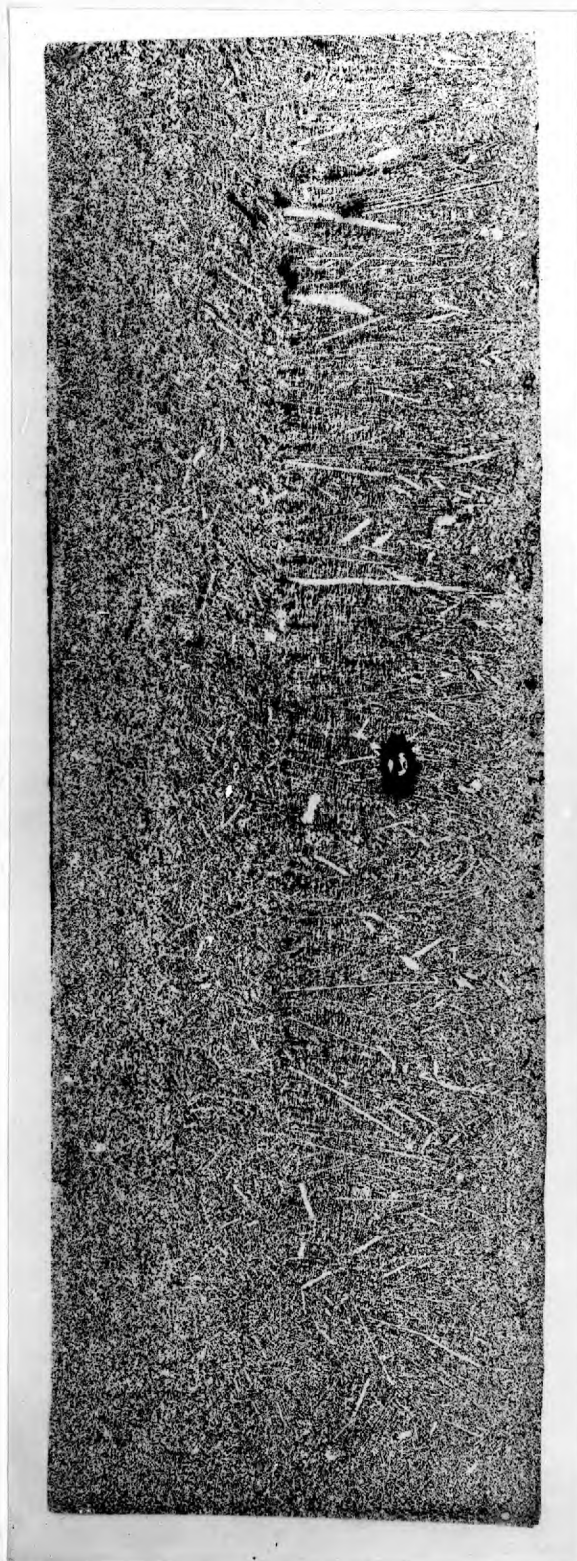


Fig. 27. Sulphur Print. Casting C2.

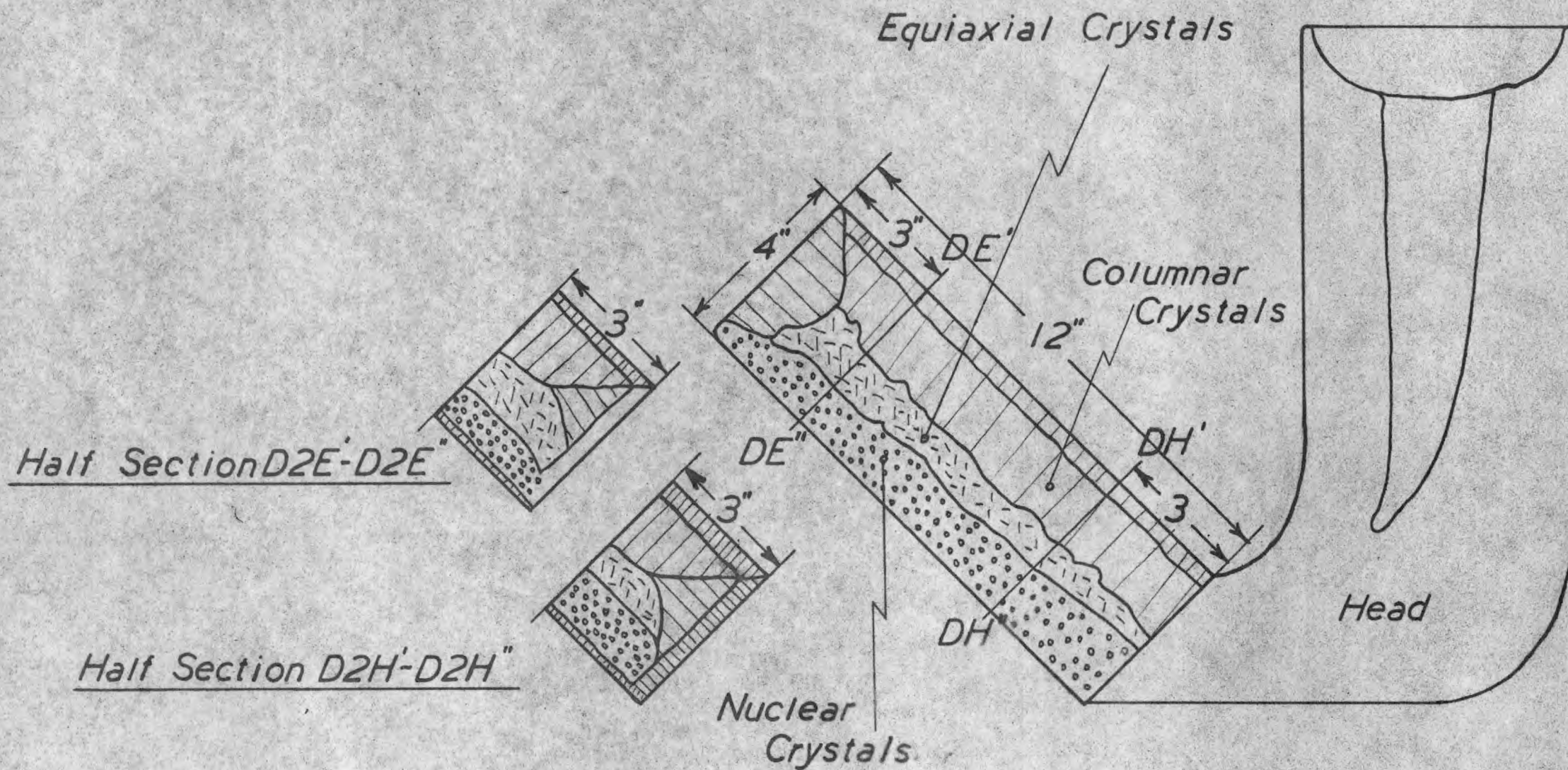
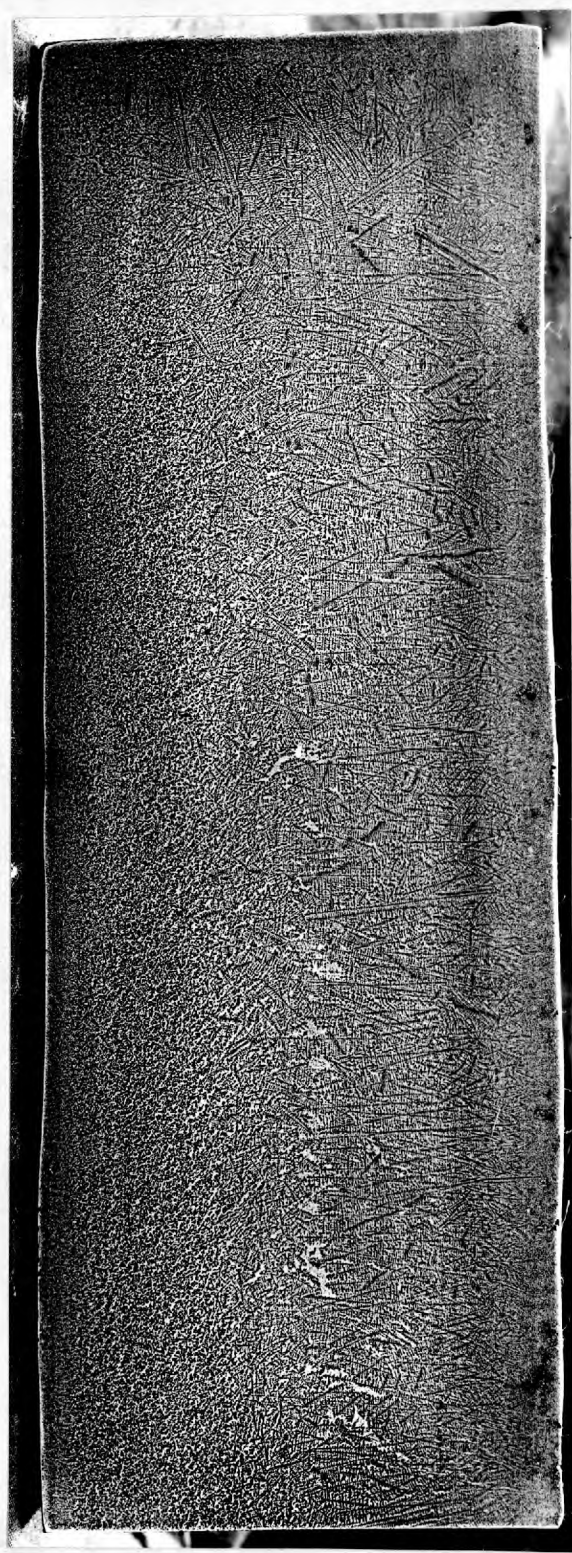
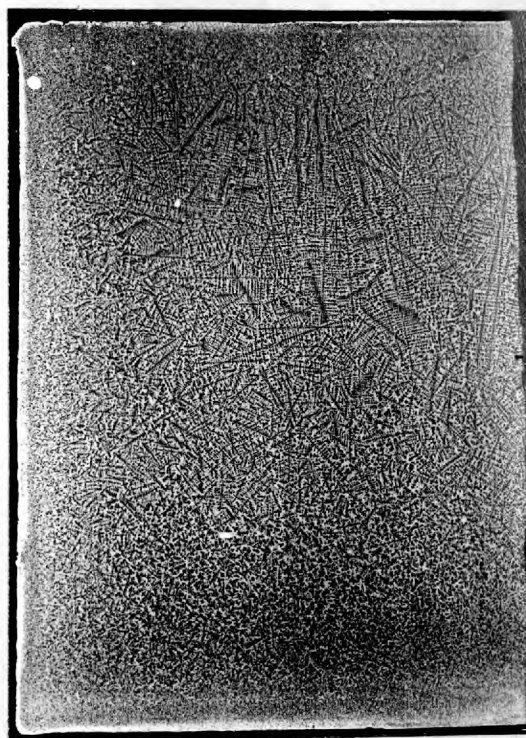


FIG.28 STRUCTURE OF CASTING D2.



HEAD

(a) Longitudinal Section

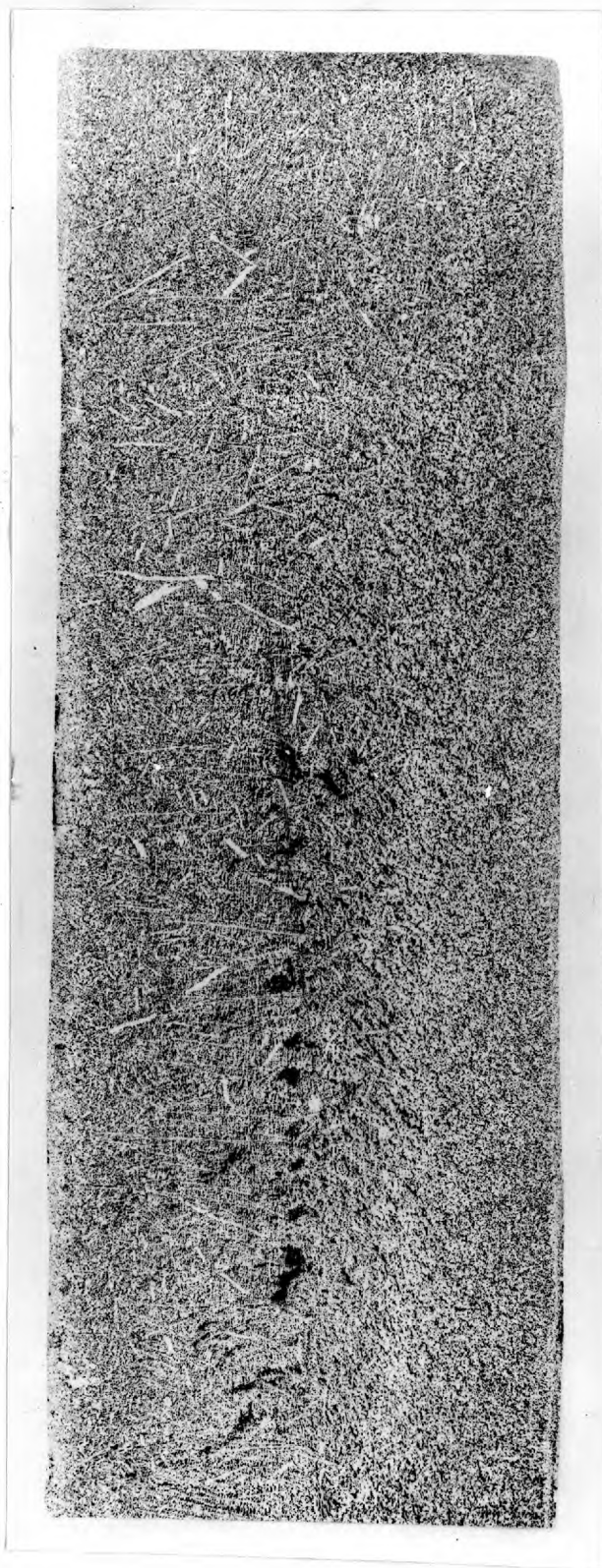


(b) Transverse Section D2E'-D2E''



(c) Transverse Section D2H'-D2H''

Fig. 29. Macrostructure of Casting D2.



HEAD

Fig. 30. Sulphur Print. Casting D2

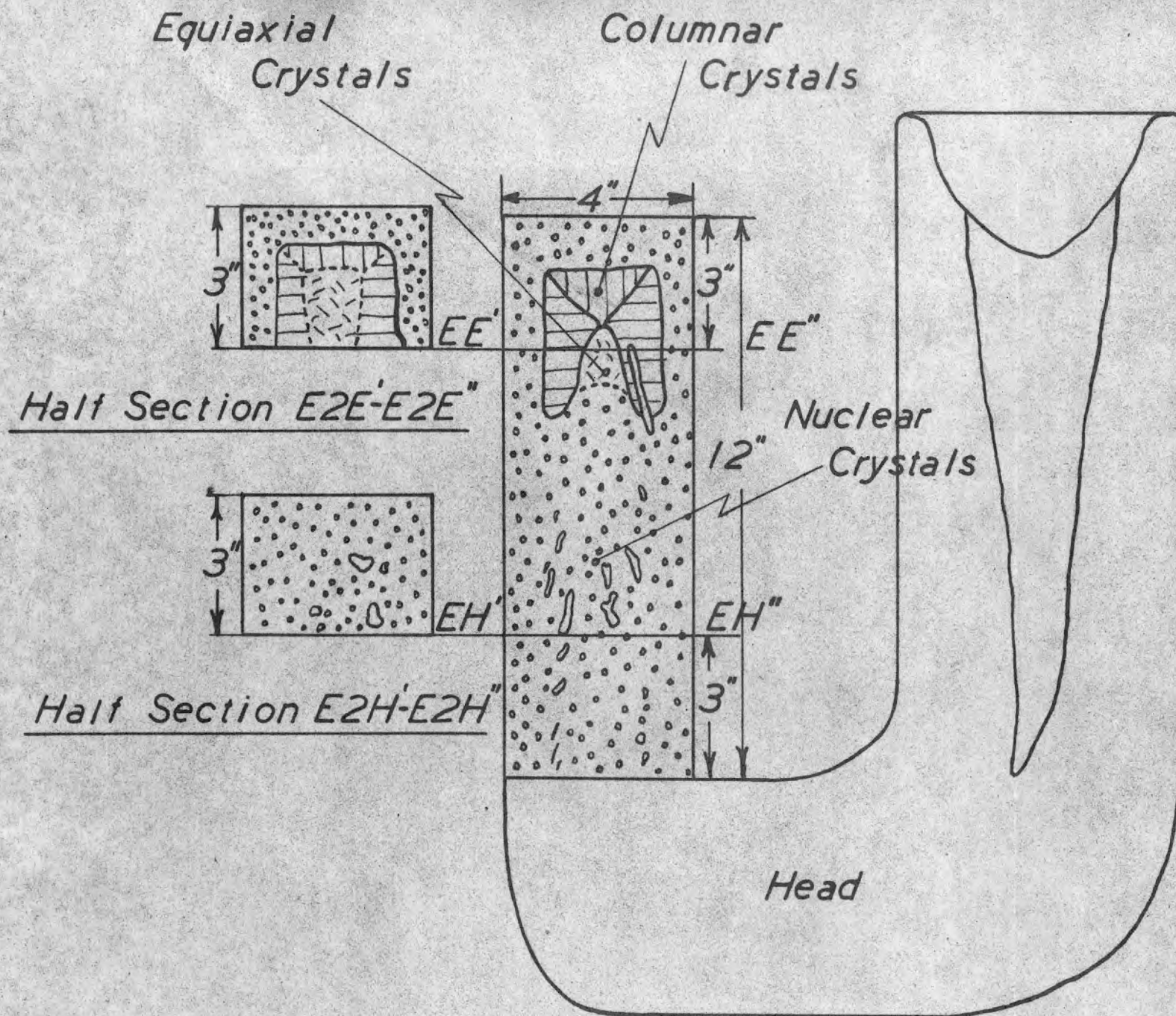
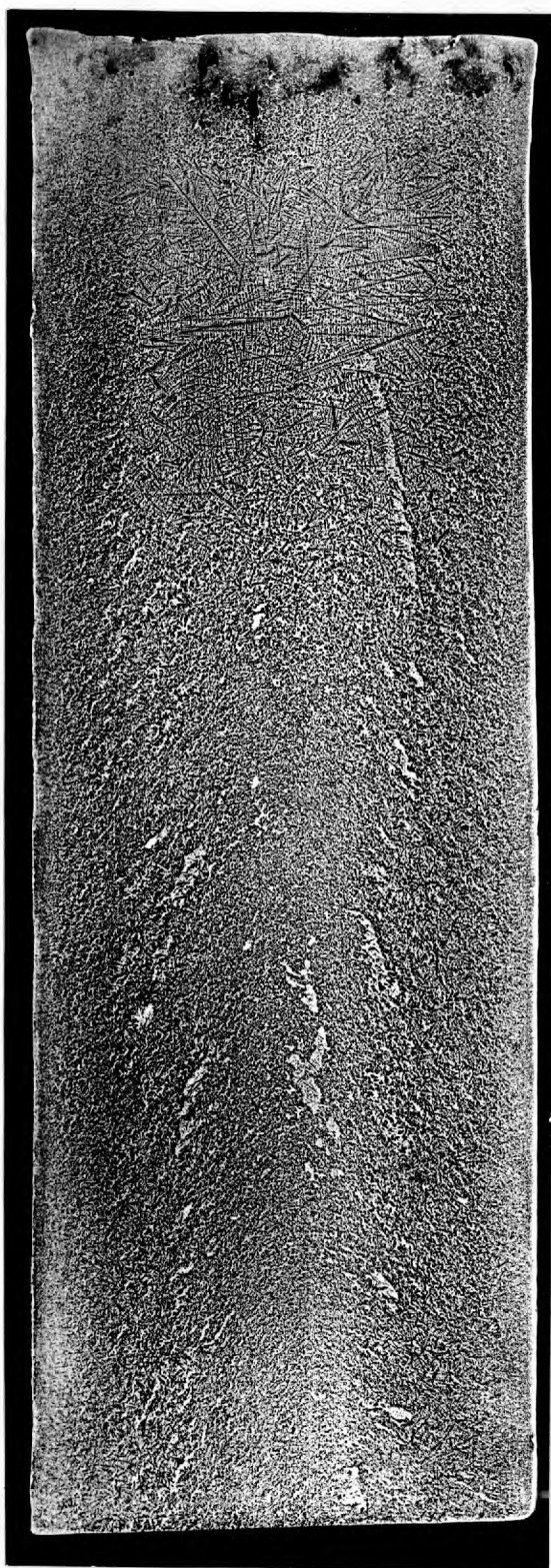
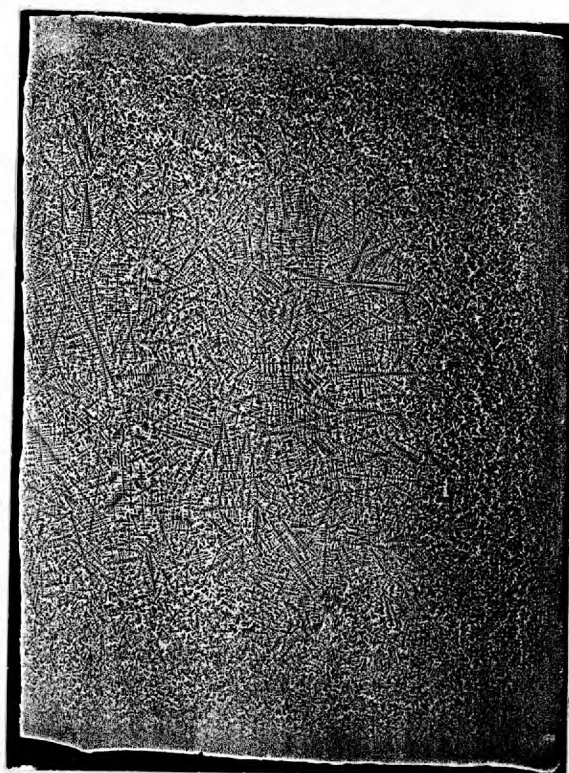


FIG. 31

STRUCTURE OF CASTING E2.



HEAD
(a) Longitudinal Section



(b) Transverse Section E2E'-E2E''



(c) Transverse Section E2H'-E2H''

Fig. 32. Macrostructure of Casting E2.



HEAD

Fig. 33. Sulphur Print. Casting E2.

SECTION THREE.

INTRODUCTION.

The present section contains an account of the investigation of certain factors affecting the solidification of small laboratory sand-cast steel blocks (measuring 5" x 3" x 2 $\frac{1}{2}$ "), the experiments being designed to add to the information obtained from the examination of the two series of larger commercial castings described in the previous section.

Firstly, a series of small blocks was cast, using feeding arrangements similar to those employed for the larger commercial series. The structures of the blocks were examined and compared with those of large castings, thus throwing some light on the effect of the size factor on the resulting type of crystallization.

Secondly, the effect of casting temperature on the crystal structure of a series of small, vertical, top-fed blocks was determined, one object being to find whether the casting temperature and the associated thermal conditions, influenced the type of crystal structure in such a manner as to account for the two forms of structure typified by castings A1 and A2 of the larger commercial series.

In the third series of experiments, the temperature conditions existing during the solidification of the steel blocks were measured. Thermocouples were introduced into the mould cavities of the two vertical blocks (one of which was fed at the top and the other at the bottom), and continuous

measurements of temperature were recorded by a high-speed recorder during the solidification period. With these results it was possible to correlate the temperature distribution within the blocks with the corresponding crystal structures.

Finally, a series of "pour-out" experiments was made, using the top-fed vertical type of block, the object being to correlate the depth of freezing with the recorded temperature measurements.

EXPERIMENTAL TECHNIQUE.

It was found that for the investigation of the factors affecting the mechanism of solidification of the laboratory sand-castings, the same basic mould design and casting procedure could be employed in all cases; the temperature measurements and "pour-out" experiments necessitated only minor modifications. Consequently, a general description of the moulding arrangements will be given now, followed by an account of the modifications that were adopted for particular purposes.

(a) Design of Castings.

The dimensions of the laboratory castings were limited by the quantity of steel that could be melted in the high-frequency furnace available, i.e. 22 lbs. The size of the block was 5" x 3" x 2½". The quantity of molten metal available for the feeder heads was not enough to ensure soundness in the bottom-fed blocks, so in making these castings collars had to be used to keep the metal in the head liquid until the blocks had solidified. Although this was regrettable, it was considered desirable that the block should be as large as possible, to avoid excessively rapid solidification. This was especially necessary with temperature determinations, as there was always an initial thermal lag due to the mass of the thermo-couple sheaths.

The designs finally chosen are shown in Figs. 34 - 38. The blocks were arranged with the 5" x 2½"

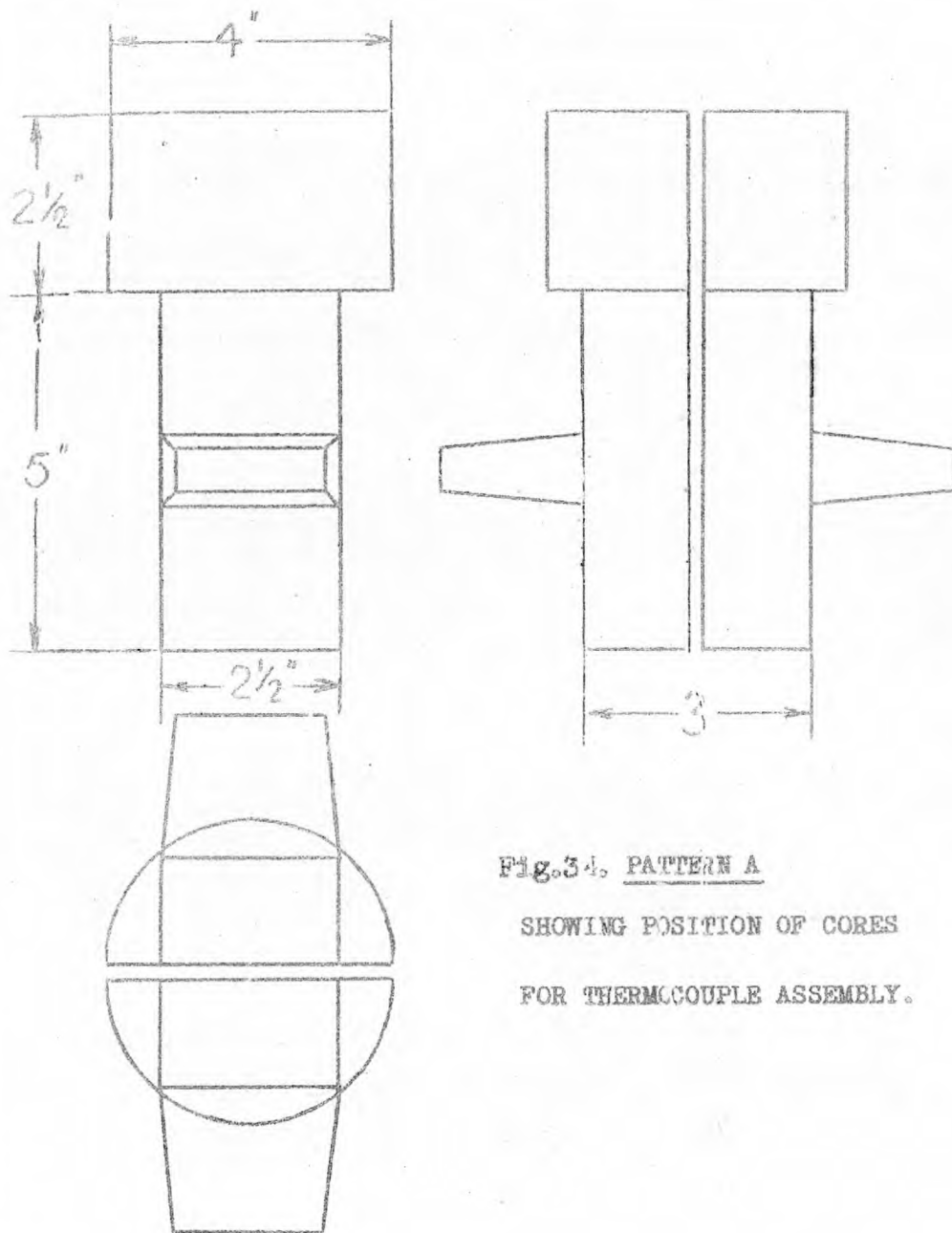


Fig. 34. PATTERN A
 SHOWING POSITION OF CORES
 FOR THERMOCOUPLE ASSEMBLY.

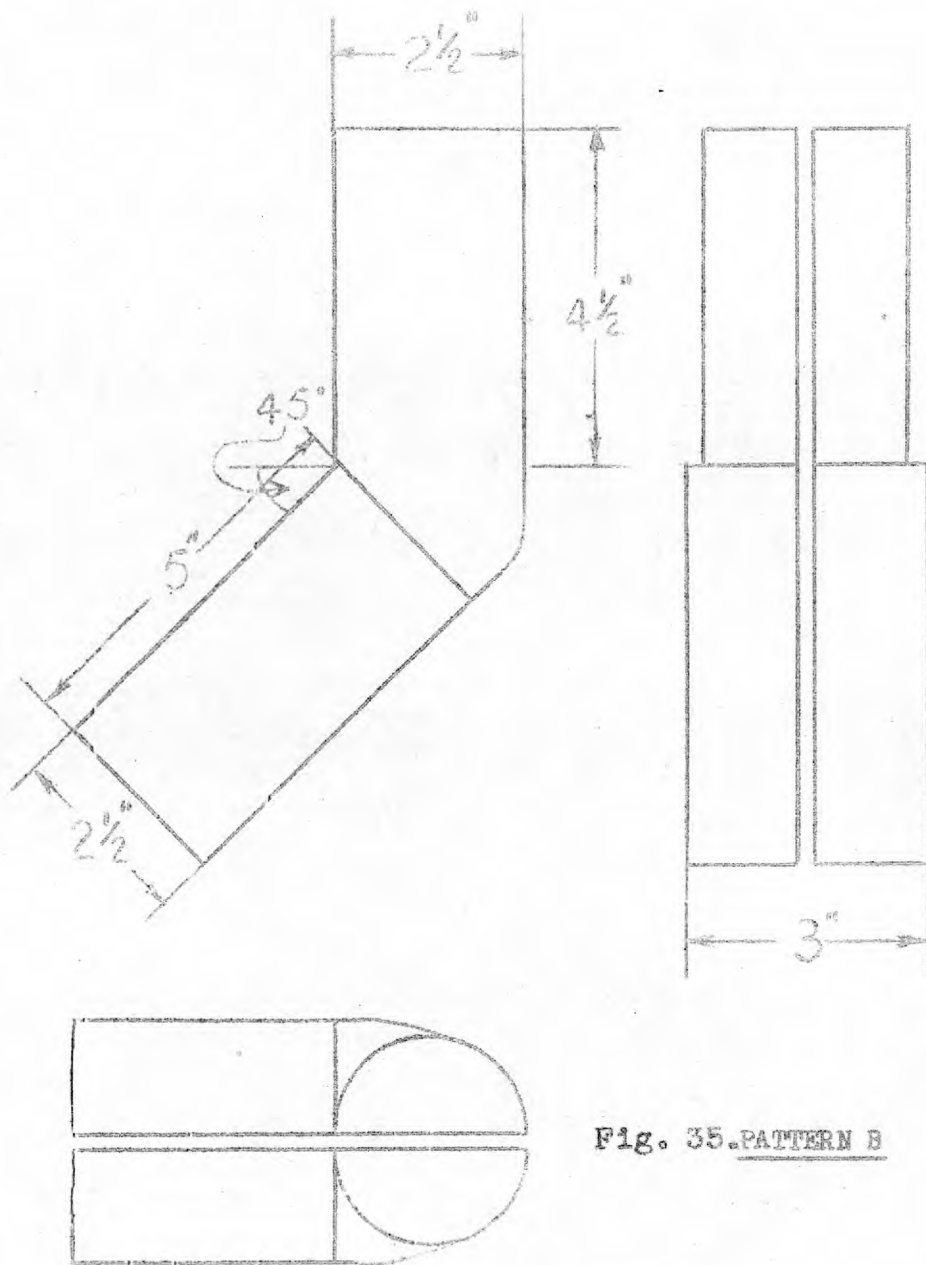
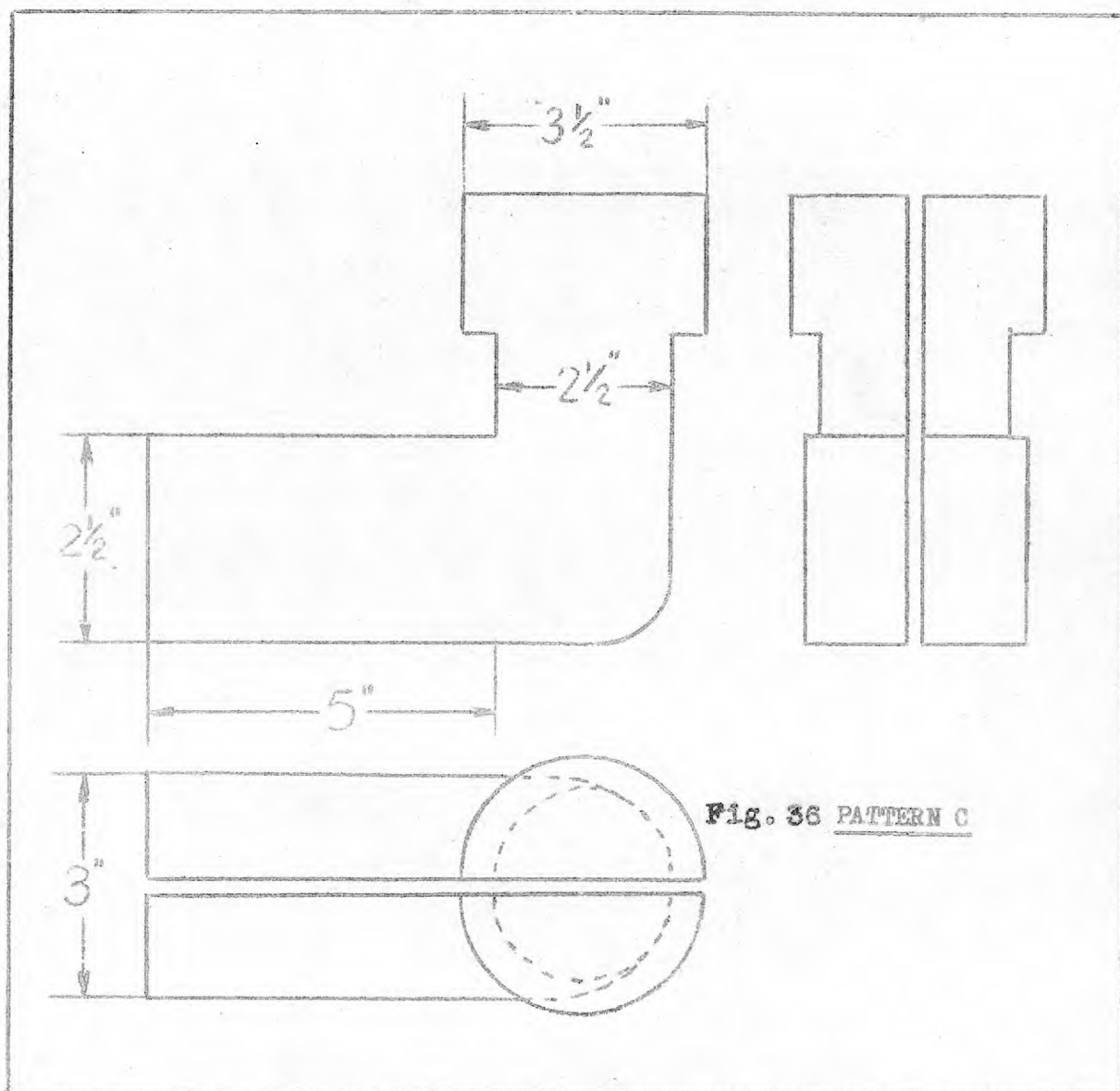
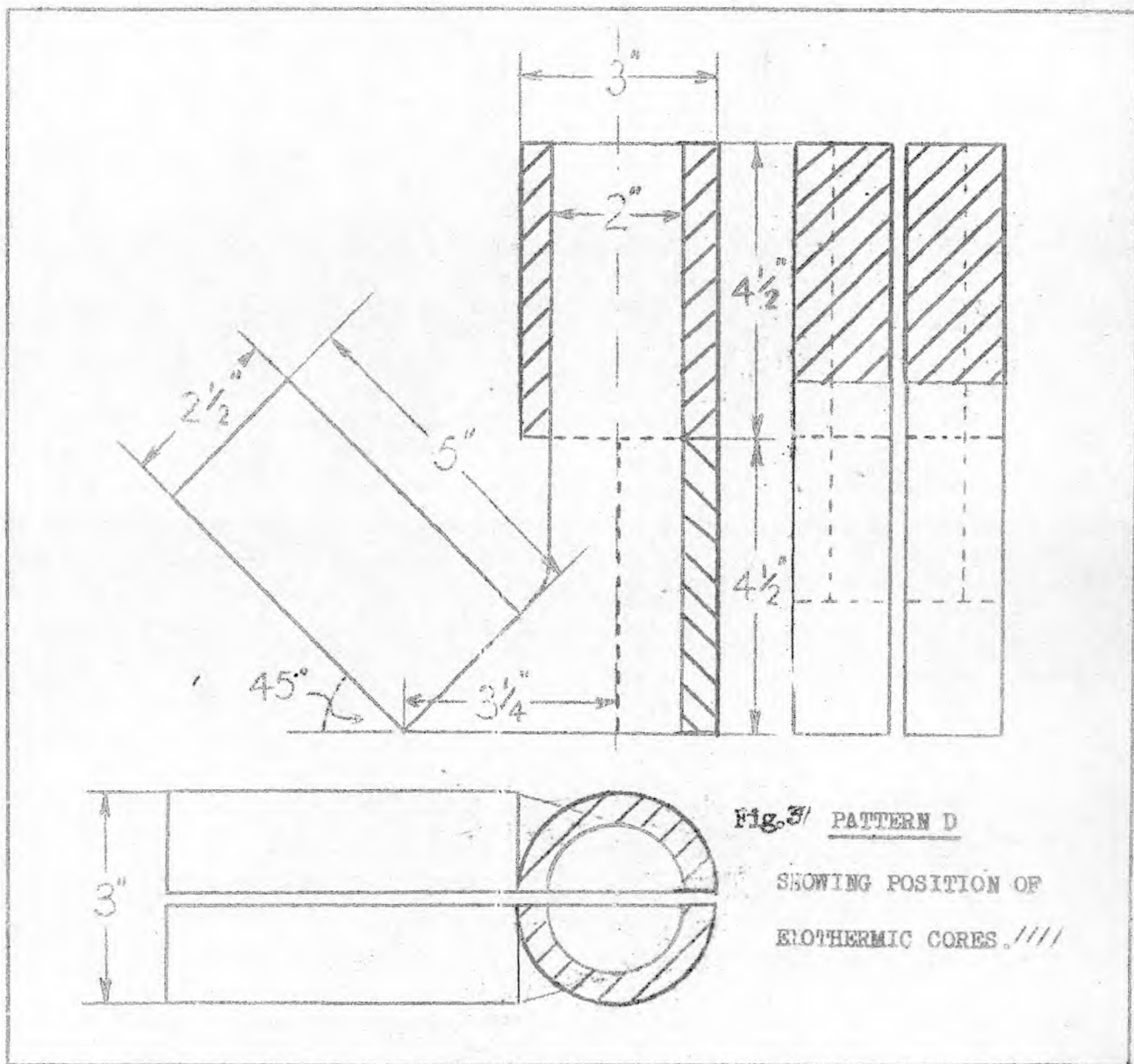


Fig. 35. PATTERN B





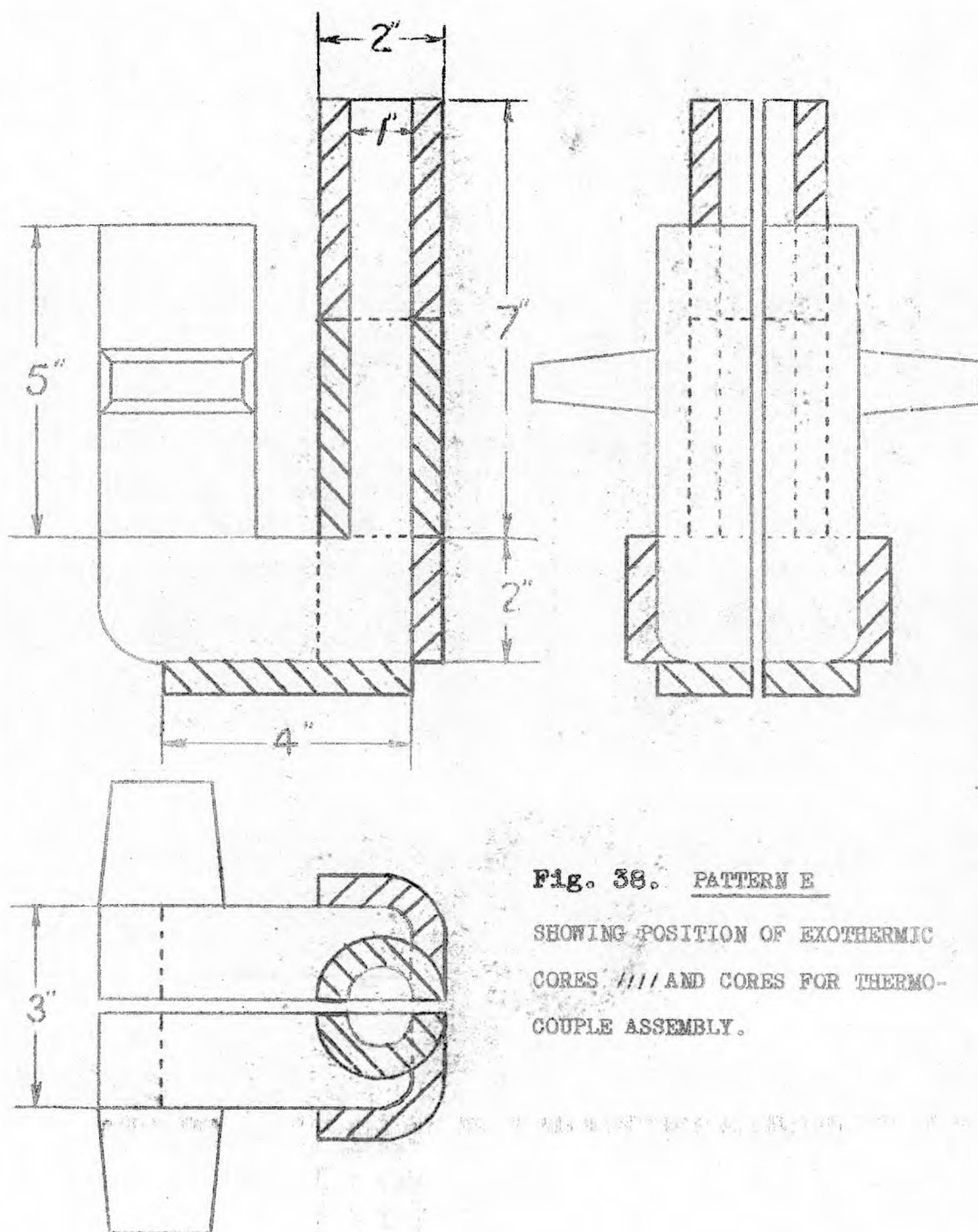


Fig. 38. PATTERN E

SHOWING POSITION OF EXOTHERMIC
CORES /// AND CORES FOR THERMO-
COUPLE ASSEMBLY.

plane vertical, in order to eliminate any side wall effects.

Casting A is a top fed vertical block, 5" x 3" x 2½", surmounted by a cylindrical head (Fig. 34).

Casting B is a top fed block inclined at 45° to the vertical. (Fig. 35).

Casting C is a horizontal block fed at one end. (Fig. 36).

Casting D is a bottom fed block inclined at 45° to the vertical. (Fig. 37). Half inch thick exothermic sleeves are placed around the head as is shown in the diagram.

Casting E is a bottom fed vertical block (Fig. 38). It contains half inch thick exothermic cores placed along the base and the vertical portion of the head.

In all cases the patterns were divided lengthwise, the section being taken half way across the 3" face.

The sizes of the exothermic cores employed in casting D and E were determined by "trial and error."

In order to have the largest feeder heads possible the castings were poured directly through the heads, and not through ingates as in normal practice.

(b) Moulding Technique.

As it was intended to measure the temperature at points within the castings during solidification, it was necessary to employ a method of moulding by which a number of thermocouples could be introduced, if required, into the mould cavities. It was also essential that the thermocouples should be accurately and firmly positioned within the cavities and yet not interfere with the normal

design of the casting.

For these reasons a horizontal snap-flask moulding technique was considered most suitable, employing patterns which were divided lengthwise, as shown in Figs. 34 - 38. This method was adopted for all the laboratory castings, whether the object was to determine temperature conditions, or to find the effects of casting temperature, or to make "pour-out" tests.

The mould containing the divided pattern was rammed horizontally in a wooden snap-flask in the normal manner. The two halves of the pattern were initially positioned on the moulding board in such a way that the position of the pattern corresponding to the head lay in contact with, and at right angles to, the side of the flask. (The flasks were originally designed so that the patterns could be interchanged as required, and were surrounded by a minimum depth of 3" sand). After separating the flasks, the patterns were carefully withdrawn; the mould was then vented and surface dried. The necessary exothermic cores for castings D and E were prepared from commercial Fosceco Feedex X16; these were baked at 200°C and given a heavy silica wash to prevent contamination of the casting with aluminium particles. A half mould, slightly modified for temperature determinations, is shown complete with exothermic cores in Fig. 39.



Fig. 39. View of Open Half Mould with Exothermic Cores.

(c) Mould Assembly.

As an illustration of the general procedure employed for mould assembly, a detailed description is given of the closing and assembling of moulds containing exothermic cores. The modifications made for temperature determinations are described later on page 80 , under "Temperature Measurement".

The drag, containing the surface-dried sand mould, was laid horizontally, and the exothermic cores placed in their appropriate positions in the mould. The cores for the cope were positioned on top of the corresponding cores in the drag. Any small pieces of loose sand were removed and the cope lowered gently into position over the drag. When satisfactorily closed the snap-flasks were removed, leaving the sand mould exposed and the mould cavity

visible. A wooden casting box was placed around the sand and clamped, although leaving the feeding-head open, cf. Fig. 40.

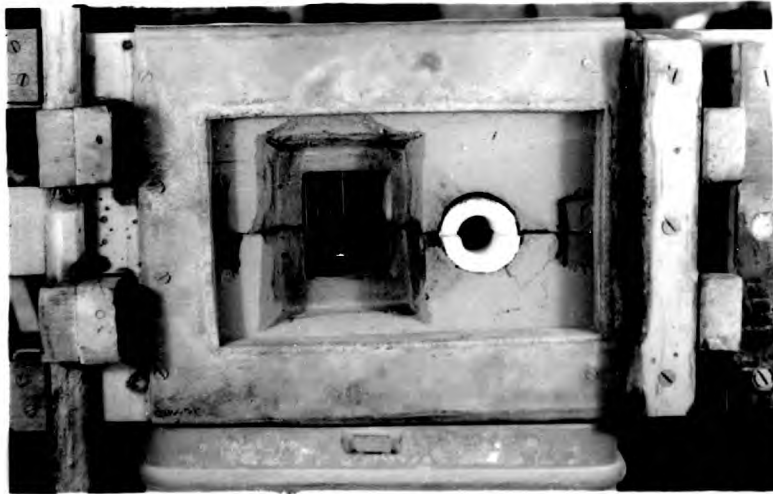


Fig. 40. Top view of open Mould showing the Head and position of Thermocouple Sheaths.

The assembly was then slowly raised into a vertical position, with the feeder head upper-most. Side plates were placed alongside the mould and clamped securely in position with clamping rings. A small sand dozzle head was then placed over the feeding head to prevent spilling of molten metal into the casting box. Figs. 41, 42 and 43 show the final arrangements for temperature measurements. In the case of "straight" casts, a similar arrangement was employed without the introduction of thermocouples, cold junctions etc.



Fig. 41. Side view of Mould Assembly
showing Thermocouple connections.

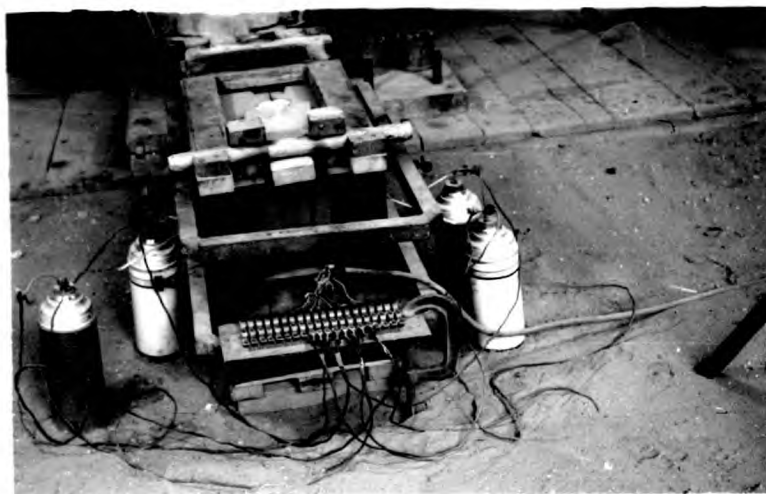


Fig. 42. End View of Mould Assembly
showing Thermocouple Connections.

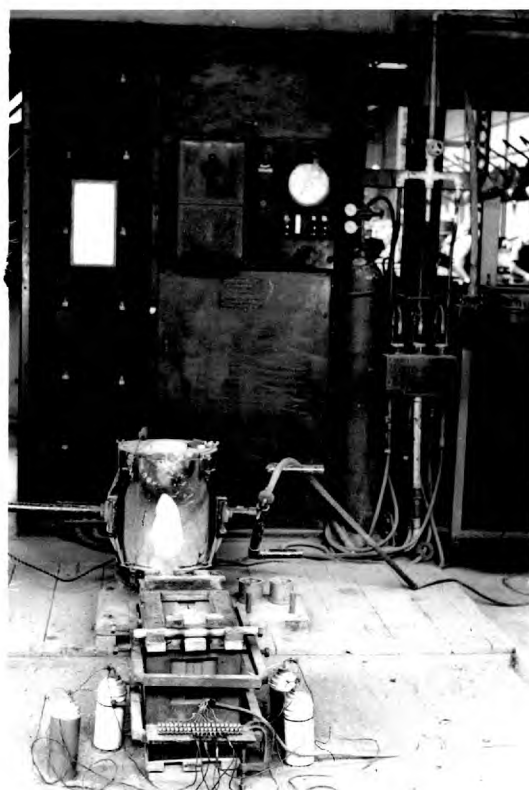


Fig. 43. General View of Mould Assembly and Furnace.

(d) Temperature Measurement.

As the purpose of the temperature determinations was to measure small differences in temperature existing during the solidification of a casting, it was essential that an accurate recording apparatus should be employed. For maximum sensitivity, it would be advantageous to use unprotected thermocouple junctions. West⁽¹⁶⁾ found, however, that in similar experiments, exposed couples rapidly failed. As the thermocouples had to be enclosed in silica sheaths, it was essential that both they and the sheaths should be as thin as possible to minimise errors due to their mass. Furthermore, the thermocouple junctions had to be capable

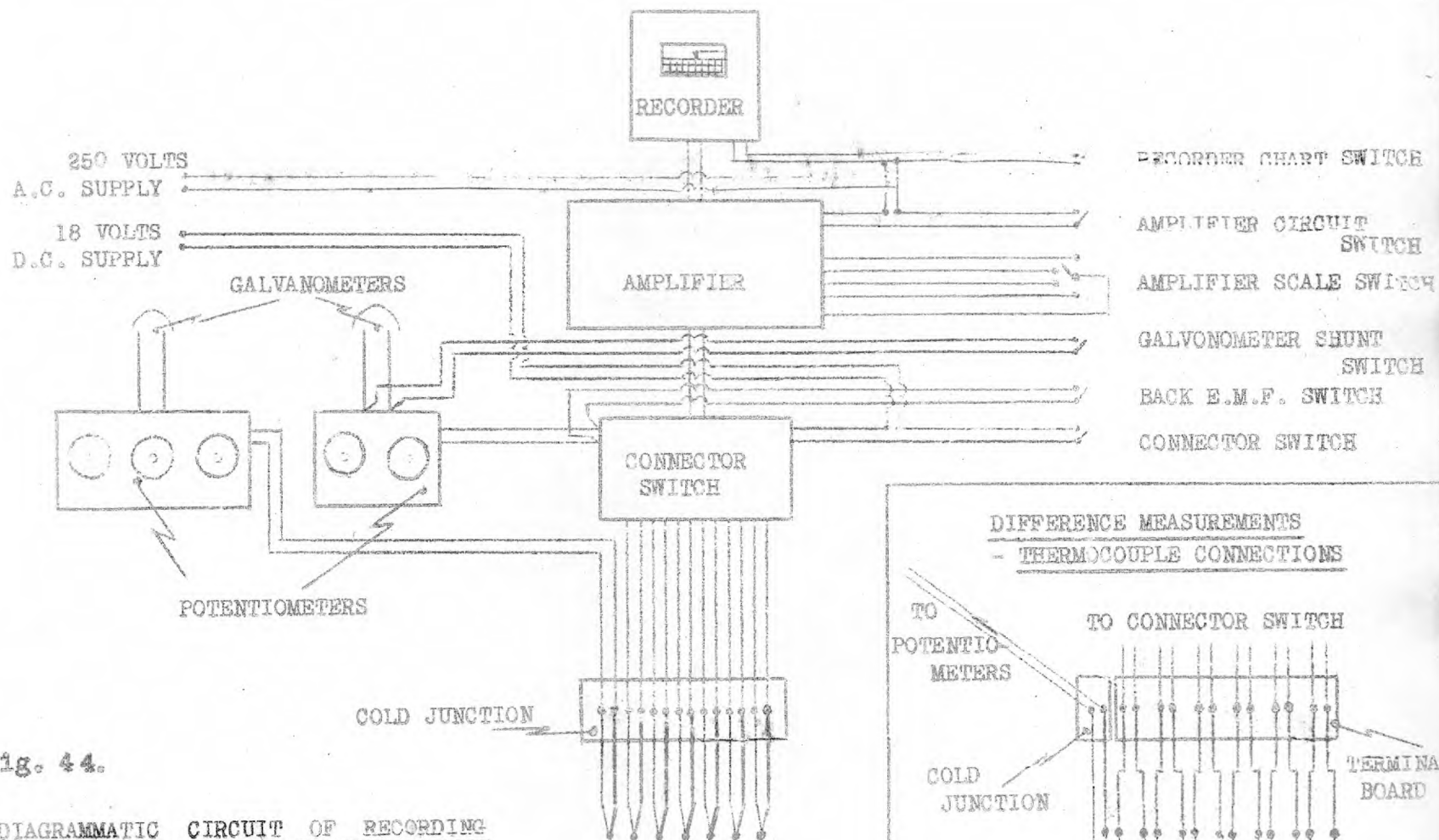


Fig. 44.

DIAGRAMMATIC CIRCUIT OF RECORDING

APPARATUS EMPLOYED FOR DIRECT TEMPERATURE MEASUREMENTS.

of being accurately positioned, and the couples had to be sensitive and reliable in indicating accurately rapid changes of temperature in the neighbourhood of the melting point of steel (1500°C). In addition, the apparatus had to be capable of recording rapid changes of temperature in quick succession from different points in the solidifying casting.

The system adopted was to employ platinum/platinum rhodium thermocouples in conjunction with a high-speed temperature recorder and an automatic selector switch. A diagrammatic circuit of the apparatus employed is shown in Fig. 44.

The actual measurement of the temperature conditions was carried out by two methods, viz:-

(i) Differential Temperature Measurement and (ii) Direct Temperature Measurement.

(i) Differential Temperature Measurement.

The differential method consisted of measuring the difference in temperature between successive positions within a casting. Consequently, if the temperature of one position was also determined on an "absolute" scale, the temperature of all the other positions could be calculated. By employing this method, it was hoped that greater sensitivity would be obtained than by the direct method, in which the actual temperature at each position is determined. However, the differential method was found to be unsatisfactory

as, in calculating the "absolute" temperatures from the differences, the errors were cumulative and the wide scatter in the results offset any advantage the method might possess. Nevertheless, the curves of temperature difference obtained showed several interesting features, and some of them ^{are} included in the thesis (Figs. 72 - 75).

The measurement of the temperature differences can be understood by considering Fig. 45.

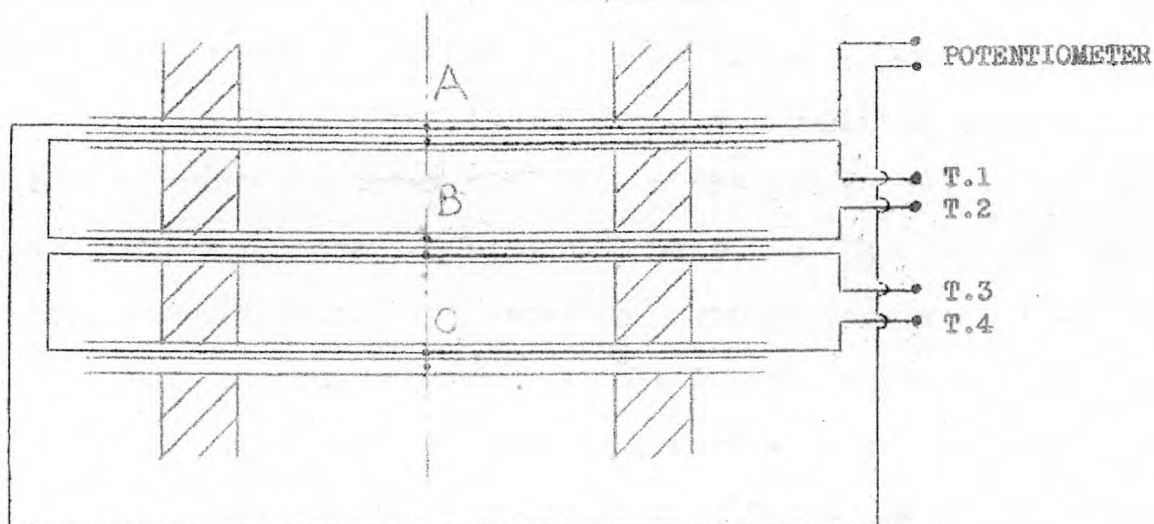


Fig. 45. Thermocouple Connections for Difference Measurements.

Consider three positions A, B and C. At each position two thermocouple junctions are situated, enclosed in twin-bore silica tubing. By connecting one thermocouple at position A to a potentiometer the "absolute" temperature at that position can be measured. If, now, the platinum wire of the other thermocouple at A is common to one at B, and the two corresponding platinum/rhodium wires T_1 and T_2 are

connected to a recorder, a measure of the temperature difference between A and B is obtained. Similarly, if a platinum wire is common between B and C then by connecting T_3 and T_4 to the recorder the temperature difference between B and C is obtained, and similarly for any other number of positions. The temperature of position C can be obtained if required, by adding the sum of the differences between A and B, and B and C, to that measured on the potentiometer connected to A, and similarly for any other position in the block. This is also illustrated in Fig. 44.

In practice it was found desirable to place the thermocouple connected directly to the potentiometer in the portion of the casting last to freeze. In this way, if the thermocouple pairs were taken in sequence away from that position, placing the final couple in the position expected to be the first to freeze, then a series of differences would be obtained which could be considered all positive or all negative. This is because, for example, if A were located in the position last to freeze, B would be always equal to or cooler than A, C always equal to or cooler than B, D always equal to or cooler than C, etc. Consequently, where the temperature distribution was being measured along the axis of a vertical, top-fed block then the couple connected directly to the potentiometer was that nearest the head and, where the distribution was being measured across the block then the couple connected directly to the potentiometer was

the one nearest to the central axis.

(ii) Direct Temperature Measurement.

The direct method consisted of measuring the temperature directly on the recording apparatus from a series of thermocouples introduced into the mould cavity. The circuit employed is illustrated in Fig.44 (facing p.71.)

The e.m.f.s. produced by the thermocouples were fed into a Tinsley D.C. high speed amplifier, via a relay-operated connector switch, and recorded by a continuous chart recorder.

The amplifying unit possessed three ranges, giving a full-scale deflection on the recorder (of 4") for an input of two, five and ten millivolts at twenty milliamperes. These values correspond approximately to 140°C, 350°C and 700°C respectively when employing the platinum/platinum-rhodium thermocouples. By replacing the galvanometer in the amplification unit by a more sensitive instrument a full scale deflection of the recorder could be obtained for an input of 0.5 millivolts. This range, however, could not be successfully used, as vibration in the foundry was sufficient to disturb the sensitive galvanometer suspension. In order to obtain the greatest accuracy for both the differential and direct method it was necessary to employ the two millivolt scale. In the case of the direct method, any e.m.f. produced by the thermocouples in excess of two millivolts was opposed

by a "back e.m.f." from an external source such as a Tinsley Vernier Potentiometer. The back e.m.f. could be varied as required. When the differential method was employed a "zero" was set on the recorder by applying a known back e.m.f. to the instrument. With a back e.m.f. of one millivolt, a trace was obtained half way across the recorder scale, and consequently any temperature differences (positive or negative) indicated by the thermocouples situated within the solidifying casting, resulted in a measureable variation from this mean position. (See Fig.46).

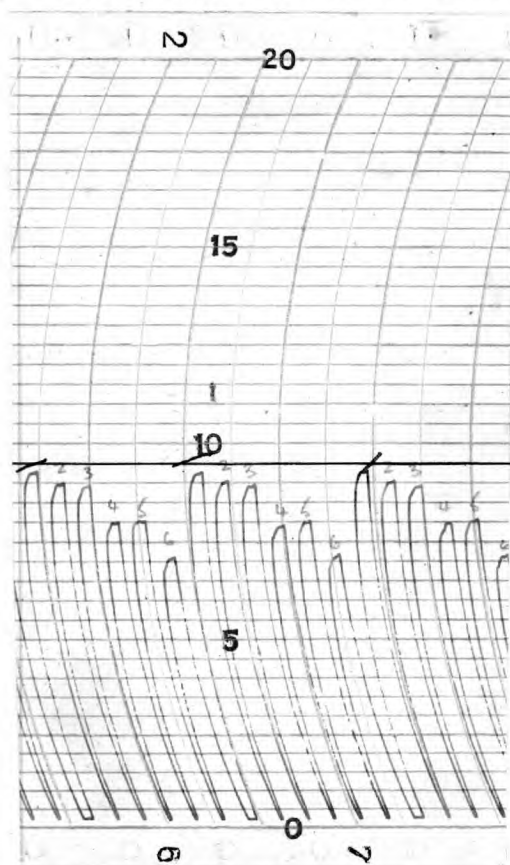


Fig.46.

Section of Recorder
Chart showing typical
pen trace for
Temperature Difference
Measurements.

By suitable adjustment of the back e.m.f. it was possible to keep the traces of the temperature differences

within the scale of the chart even when the differences in temperature between two positions in the block were at a maximum. It was possible to use the two millivolt scale in all measurements except those taken along the axis of the blocks, where, after a certain time of cooling, increasing temperature gradients made it necessary to revert to the five millivolt full-scale deflection (i.e. approximately 350°C).

By employing a relay-operated selector switch as many as eighteen thermocouples could be used in any one cast. The number normally employed, both for the differential and the direct temperature measurements, varied between four and seven. If each couple was connected to the recorder for approximately 1 second, it was found that by employing a chart speed of seven and a half inches per minute, there was sufficient time for the pen attached to the recorder to reach its maximum position and for a trace of suitable length to be made before the connection was switched to the next thermocouple. (The speed of rotation of the switch could be controlled by placing suitable condensers across the relay circuit.) A typical series of pen traces, showing direct temperature measurements, is shown in Fig. 47.

The recorder had a response time of approximately one-fifth of a second.

The thermocouples employed both in the differential and direct temperature measurements were made

from 0.2 m.m. diameter platinum/platinum-rhodium wire.

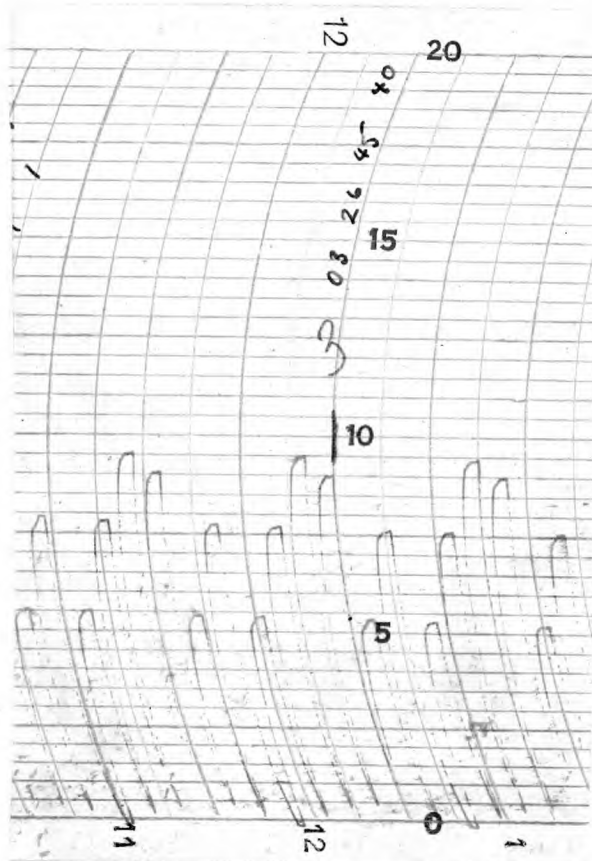


Fig. 47.

Section of Recorder
Chart showing typical
pen trace for direct
Temperature
Measurements.

This was guaranteed by the manufacturers to $\pm 2.0^{\circ}\text{C}$ in the temperature range employed ($1400^{\circ}\text{C} - 1600^{\circ}\text{C}$). Therefore, as the accuracy of the recorder was claimed to be 2% - which corresponds to 3°C in the two millivolt range - it could be considered that the total error in the readings would be of the order of $\pm 5^{\circ}\text{C}$. The "cumulative" error arising in estimating "absolute" temperatures from the results obtained in the differential temperature measurements could be much higher than this. Experience with the instrument has shown, however, that by applying certain "standardizing" procedures during the direct temperature experiments, much of the error could be eliminated.

Although the temperature indicated by a given couple might be 5°C from the true temperature a consideration of the sources of error shows that this would remain more or less constant during an experiment:

- Source 1. Variation in the characteristics of the couple wires "as received" and during use. All the wires were taken from the same batch and to minimise the latter effect, fresh junctions were fashioned before each experiment was made.
- Source 2. Errors arising in the connector switch. This error is probably constant for each set of contacts and remains substantially constant during an experiment.
- Source 3. Error in the recorder; also substantially constant during an experiment.

Hence, by "standardizing" immediately after the completion of each experiment, much of the error inherent in the instruments was eliminated; but not that inherent in the thermo-couple wires. The "standardizing" operation consisted of "shorting" the thermo-couple at the terminal board and passing standard e.m.f.s. through the recorder (0.25, 0.50, 0.75, 1.00, 1.25 and 1.50 millivolts). Thus the true values represented by the traces could be computed. This procedure, as well as allowing for any variation in the error in the amplifier and connector switch between each experiment, was especially necessary for determining the error in the recorder. (It was found that the recorder pen did not always return to the same position after each experiment. This was due to the method of suspension of the pen.)

By assuming the liquidus arrest temperature to be the same for all positions in a given casting, it was possible to make a correction for the variation between different thermo-couple wires and for certain errors arising in the apparatus. Evidence for the validity of this assumption was found by direct experiment. Several repeat experiments were made for one type of casting to test whether the apparent differences of a few degrees in the arrest temperature were real or an error attributable to the apparatus and the thermocouple wires. It was found that these differences were not reproducible, and they were therefore ascribed to experimental error.

Using the two millivolt range and the methods of correction described above, it was considered that a real temperature variation of $\pm 0.5^{\circ}\text{C}$ could be determined. (Although this method of correction is described primarily for direct temperature measurement it is possible, to a certain extent, to correct some of the errors arising in the differential method. However, in this case there are other sources of error such as the varying length and temperature of the common thermocouple wire and the absence of a cold junction, with the result that the possible "cumulative" error is excessively high.

In this investigation the International Temperature Scale of Celsius has been adopted. This is approximately the same as the Centigrade scale, differing only in a constant term in the temperature function. On the Celsius scale the lower values of temperature are the same as for Centigrade, but the higher values differ slightly. For example, 1500° Centigrade is equivalent to 1497.8° Celsius.

(e) Modifications to the Moulding Technique necessitated by Temperature Determinations.

It has been mentioned that one of the most important considerations in determining the temperature distribution in the solidifying castings was to ensure that the thermocouple junctions could be accurately positioned in the mould cavity and that they should not "float" or wander during the solidification process. This was attained by employing a thermocouple-sheath core assembly which could be introduced complete with positioned thermocouple wires into the mould cavity at a suitable stage during the assembling of the mould.

The sheath assembly employed for direct temperature measurements across a block is shown in Fig. 48. The core assembly consists of two wedge shaped sand cores carrying accurately spaced single-bore silica tubing. The sand cores were made from glycol-bonded core sand, rammed in a core box and finally hardened and strengthened for handling by baking at 150°C . A similar

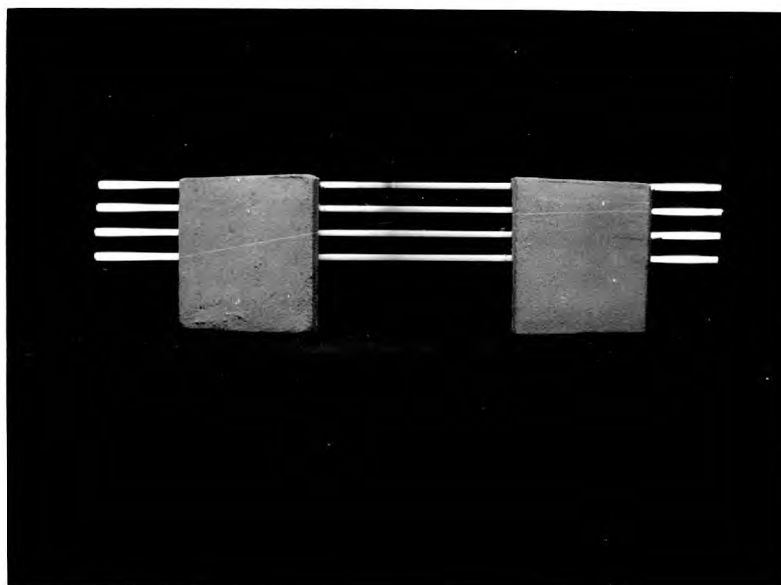


Fig. 48. Thermocouple Sheath Assembly.

though larger form was employed for temperature measurements along the axis of the block. In the case of the differential method twin-bore silica tubing was used in place of the single-bore type. The dimensions of silica tubing found most satisfactory were:-

- (a) Single-bore tubing: Bore 0.04" : External Dimension 0.1".
- (b) Twin-bore tubing (Oval section) : Bore 0.02"
External Dimensions 0.125" x 0.0885"

During a series of preliminary experiments considerable difficulty was experienced in obtaining satisfactory results. This was due to a variety of causes, and although the measurements were being determined by the differential method it was concluded that similar difficulties would have arisen employing the direct method. These difficulties are briefly discussed below, followed by a

description of the procedure finally adopted.

In order to position the thermocouple sheath assembly accurately in the mould cavity, two wedges of similar size to those forming part of the assembly were added to the pattern at the required position. (Figs. 34 and 38) facing page 66. Consequently after the patterns were removed from the mould, two cavities remained into which the assembly could be placed. Originally, thin slots were cut through the mould walls, both in the drag and the cope, to enable the silica sheaths to pass through to the outside of the mould. The sheath assembly was then placed in position in the drag before the two halves of the mould were placed together, after which the mould was assembled ready for pouring in the manner described earlier on Page 68. It was frequently found, however, that the silica sheaths were sheared, due to slight movement between the two halves of the mould during the assembling operation. Wooden spacers placed between the two cores of the sheath assembly failed to prevent the breakage.

In order to avoid breakage it was therefore essential that the sheath assembly should be introduced into the mould after it had been assembled, when no further stress would be imposed on the silica sheaths. This was accomplished in the following manner:-

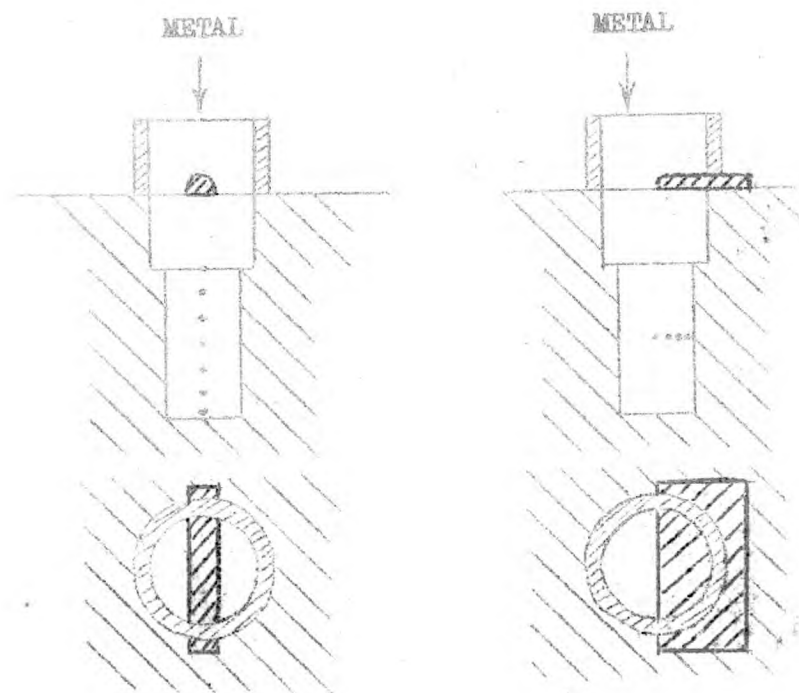
The narrow slot cut in the cope to allow the silica sheaths to extend through the mould wall,

was enlarged in size sufficiently to allow the whole sheath assembly to be introduced through it. The mould was then assembled in the normal manner and was raised vertically. The sheath assembly slid into position through the hole in the cope ~~within~~, the sand cores thus becoming part of the mould wall (Fig. 40 p.69). The remainder of the cavity on the outer side of the cope was carefully filled with green sand, thus leaving the silica sheaths projecting through the mould walls. The side plates were placed in position, clamped securely with clamping rings, and the completed assembly was placed in front of the furnace. (Figs. 41, 42, 43. p.70).

This method was found to be entirely satisfactory. Misgivings that the weight of molten metal in the casting might force the sheath assembly back through the mould wall, proved to be groundless.

It was found necessary, however, in the case of the top poured blocks (Casting A) to protect the silica sheaths from the direct impact of the molten metal during pouring. This was achieved by incorporating into the dozzle head a refractory bar, or plate, covering the sheaths. (See Figs. 49 and 50, page 84).

Prior to the introduction of the sheath assembly into the mould, the thermo-couple junctions were fixed in position. The thermocouple wires were threaded through the appropriate sheath according to the type of measurement to be made (differential or direct), until the junction was mid-way between the two sand cores. The wires were held in position by wedging them against the sheaths with small pins.



Figs. 49 and 50. Guards for Thermocouple Sheaths.

The necessary connections between the thermocouples and the recording apparatus were then made as follows:-

- (1) In the case of the direct temperature measurements the thermocouple wires, insulated by twin bore silica tubing, were connected to the recording apparatus by a cold junction and a terminal board, (Figs. 41, 42, 43. p.70.) Any necessary final adjustments of the position of the thermocouple junctions were made after examination through the cavity or head at the top of the mould. (Fig.40. p.69.) Final tests were made of all the circuits, the thermocouple junctions being warmed and the apparatus run without a back e.m.f. to observe whether temperatures were being recorded.
- (2) In the case of the differential temperature measurements, the common thermocouple wire from

each pair of successive positions was insulated with silica tubing before being introduced into the mould. When in position in the mould, the corresponding pairs of like thermocouple wires were insulated with twin-bore silica tubing and connected directly to the appropriate positions on the thermocouple board and hence to the recording apparatus. Any necessary adjustments were made to the position of the thermocouple junctions and the circuits tested in a similar manner to that employed for the direct temperature measurement.

(f) Modifications to Mould Arrangement for "Pour-out" Tests.

To assist in the interpretation of the temperature measurement experiments, a series of "pour-out" tests was made in which castings were bled after certain time intervals, examination of the resultant thickness and contour of the solidified shell giving a measure of the progress of solidification. The "pour-out" tests were limited by practical considerations to the vertical casting fed at the top.

In order to achieve the "pour-out" effectively at a given time, it was essential to invert the mould completely, in the shortest possible time, and to allow any molten material to run out freely. Consequently it was necessary for the head to remain molten at the top.

The rapid inversion was obtained by the following method. Two metal struts were strapped across both ends of the casting box. One end of the pair of struts was employed as a lever, whilst to the other ends were attached arc shaped metal rockers. When the lever was lifted, the casting box,

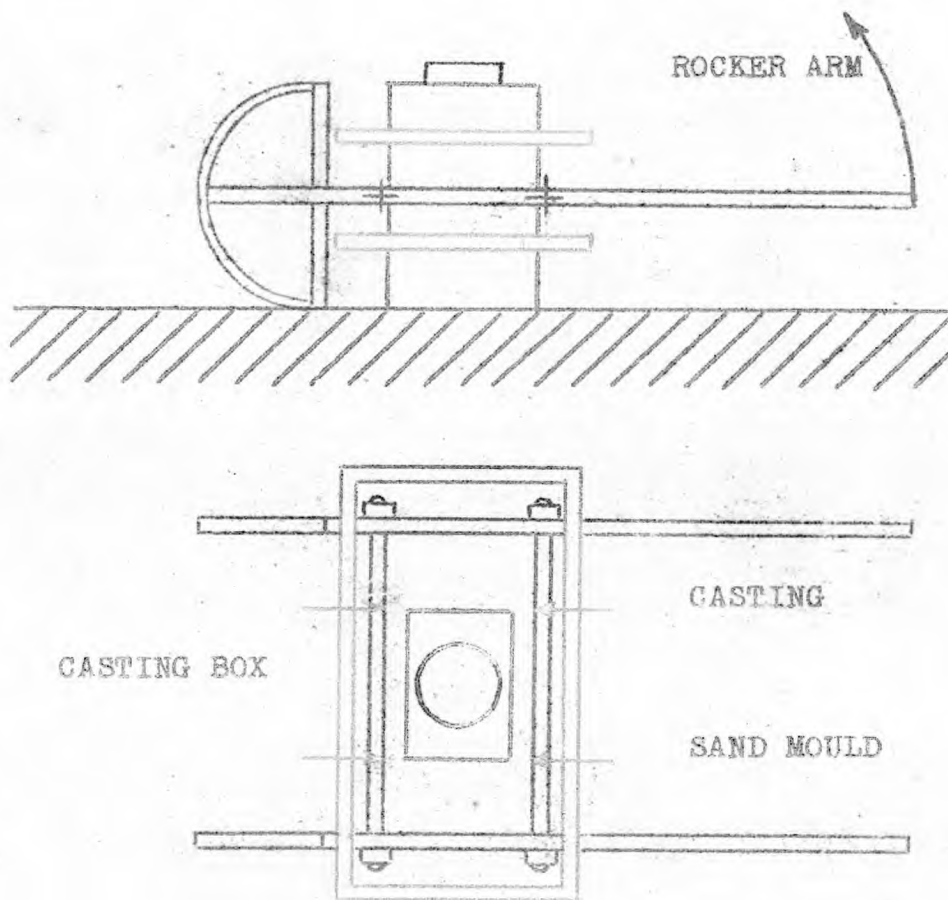


Fig. 51. MOULD INVERSION ARRANGEMENT FOR
POUR OUT TESTS.

with the casting, rotated about the arc until it was finally inverted. The whole operation took about three seconds. (See Fig.51).

The head was kept open by the addition of a small quantity of exothermic compound scattered on top of the metal after the mould had been filled. By this method it was possible to bleed the casting up to four minutes after pouring.

The mould was made and assembled in the same manner as previously described on page 67.

Melting and Casting Procedures.

The same basic melting and casting procedure was adopted for the following series of experiments which, to recapitulate, comprised the following:-

- (a) Investigation of the Effect of the Size Factor;
- (b) Investigation of the Effect of Casting Temperature;
- (c) Measurement of Temperature Conditions;
- (d) "Pour-Out" Tests.

Certain modifications had to be made to the basic procedure for the special purposes of the different sections of the investigations, so they will be described in the order given above.

(a) The Effect of the Size Factor.

For this section of the investigation, a series of five small sand castings was made under conditions similar to those under which the series of larger industrial castings had been made. That is to say, the arrangement of the feeder-heads and the "positions" of the blocks were the same, and the steel was of approximately the same chemical composition and was cast with the same superheat. Briefly, the procedure was as follows:-

In each case a straight charge of 20 lb. of "40" carbon steel rod was melted in an acid lined high-frequency induction furnace (this weight was somewhat higher than the nominal 18 lbs. capacity of the furnace). To assist in a rapid melt and to reduce carbon loss, the $\frac{1}{2}$ " diameter steel rod was cut into small lengths and stacked horizontally

in the furnace pot. A clear melt was obtained in about $1\frac{1}{2}$ hours.

It was realised that the melting operation would result in a loss of manganese and silicon, as well as carbon, but as it had been found that the chemical composition had little effect on the crystal structure it was decided not to make alloying additions: it was felt that there would be less variation in composition from cast to cast if no additions were made.

In preparation for casting, the temperature of the molten metal in the furnace was raised to within 15°C of the pouring temperature, and after removing any slag, the melt was "killed" rapidly with aluminium. It was generally found that after this operation had been completed the melt was at the required casting temperature; if necessary the temperature was adjusted. The furnace was then switched off and the metal poured directly into the mould. The temperature measurements were made with an immersion pyrometer and a Tinsley potentiometer.

After examining the first series of castings, it was found that the feeding had been insufficient in castings D and E. These were recast several times with exothermic sleeves until reasonably sound castings were obtained.

Melting and Casting Details:-

Charge: "40" Carbon Steel Rod 22 lbs.

(C 0.33%; Si 0.20%; Mn 0.76%
P 0.019%; S 0.019%;)

Deoxidizer: Aluminium Rod ($\frac{1}{8}$ " diameter) 3 lbs/ton.

Pouring Temperature: 1575°C.

<u>Casting</u>	<u>C%</u>	<u>Si%</u>	<u>Mn%</u>	<u>S%</u>	<u>P%</u>	<u>Remarks</u>	<u>Liquidus*</u>	<u>Super-heat</u>
WOX6A	0.21	0.18	0.19	0.026	0.033	Sound	1509°C	66°C
WOX7B	0.29	0.09	0.37	-	-	Sound	1500°C	75°C
WOX3C	0.29	0.08	0.28	-	-	Sound	1500°C	75°C
WOX4D	0.29	-	-	-	-	Cavity	1500°C	75°C
WOX8D	-	-	-	-	-	Cavity despite enlarged head.	Approx. 1500°C	75°C
WX2D	0.28	0.08	0.44	-	-	Sound, with exothermic sleeves.	1501°C	74°C
WOX5E	0.30	-	-	-	-	Cavity	1499°C	76°C
WX1E	-	-	-	-	-	Cavity despite exothermic sleeves.	Approx. 1500°C	75°C
WX3E	0.28	0.07	0.39	0.022	0.024	Sound, with exothermic sleeves and plate.	1501°C	74°C

* The liquidus values are taken from the equilibrium diagram (Fig.76. p.118.)

(b) The Effect of the Casting Temperature.

In this investigation a series of five small vertical top-fed sand castings was made, with superheats varying from 0°C to 125°C above the liquidus.

The melting and casting procedure was exactly the same as that described in the previous section.

In the case of the casting poured at the liquidus temperature (0°C superheat), it was found necessary to melt the metal completely at a temperature above the liquidus and then to allow it to cool to the required temperature. It was observed, in pouring the metal, that there were a number of small mushy regions in the melt. This indicated that the metal was actually at the liquidus temperature, and that crystallites were forming within the liquid.

Melting and Casting Details:

Charge: "40" Carbon Steel Rod 22 lbs.
 (C 0.33%; Si 0.20%; Mn 0.76%;
 P 0.019%; S 0.019%;)
 Deoxidizer: Aluminium Rod ($\frac{1}{8}$ " diameter) 3 lbs/ton.

<u>Casting</u>	<u>Pouring</u> <u>Temp.</u>	<u>C%</u>	<u>Si%</u>	<u>Mn%</u>	<u>P%</u>	<u>S%</u>	<u>Liquid</u> <u>-us</u>	<u>Calc.</u> <u>Super</u>	<u>Actual</u> <u>Super</u>
								<u>-heat</u>	<u>-heat</u>
SHA1	1500°C	0.28	-	-	-	-	1501°C	0°C	0
SHA2	1525	0.29	0.09	0.35	-	-	1500°C	25	25°C
SHA3	1550	0.29	-	-	-	-	1500°C	50	50°C
SHA4	1575	0.21	0.17	0.19	0.032	0.026	1509°C	75	66°C
SHA5	1600	0.32	0.08	0.35	-	-	1497°C	100	103°C
SHA6	1625	0.27	-	-	-	-	1502°C	125	123°C

(o) Temperature Measurements.

The melting and casting procedure was again essentially as described above. In all the experiments, the casting temperature was 1575°C , i.e. the superheat was approximately 75°C .

(1) Differential Method:

In preparation for the determination of the temperature distribution within the solidifying block, the amplifier circuit was switched on during the "melting down" operation. This was essential as the circuit needed a little time in which to "warm up". While the melt was being superheated, the potentiometer connected directly to the couple in the mould cavity was set at 17.8 millivolts, (corresponding to 1525°C), and the potentiometer supplying the "zero" to the recorder was set at 1.0 millivolts and its galvanometer "shorted". (This latter adjustment positioned the "zero" on the recorder half way across the recorder scale). The connector switch was set ready to pass to the first stud connecting the thermocouples to the amplifier, whilst the back e.m.f. to the amplifier was also "shorted". As soon as pouring commenced, the recorder chart was set in operation and a mark made to denote the beginning of pouring. Simultaneously a stop watch was started and the time noted when the temperature of the "direct" couple fell to 1525°C , after which an inverse rate cooling curve was determined by the normal method, until solidification of the block was

complete and it had cooled below the solidus temperature to about 1350°C .

Immediately pouring had been completed, the connector switch was set in operation and the back e.m.f. applied to the amplifier. It was found that if the thermocouples were "switched in" earlier, the temperature differences were liable to be excessively high during the first few seconds. Once the switch had been started, it was allowed to operate until solidification had been completed. The recorder was then standardised as described on page 78.

(11) Direct Method.

As before, the amplifier circuit was switched on during melting down and the final preparations were made during the superheating period. The external back e.m.f. source was set at 17 millivolts and the input to the amplifier "shorted". The selector switch was then set ready to pass to the first stud connecting the thermocouples to the amplifier. As soon as pouring commenced, the recorder chart was started and marked. Immediately the mould had filled the selector switch was set in operation and the back e.m.f. applied to the amplifier. The apparatus was left running until the casting had passed the solidus temperature; this normally took between 13 and 14 minutes. During the cooling of the block it was necessary to adjust the back e.m.f. by steps of 0.5 millivolts, to keep the readings on the recorder scale. When solidification had been completed, the recorder was

standardized as has been previously described.

Melting and Casting Details.

Charge: "40" Carbon Steel Rod 22 lbs

(C 0.38%; Si 0.17%; Mn 0.79%;
P 0.031%; S 0.021%;)

Deoxidizer: Aluminium Rod ($\frac{1}{8}$ " diameter) 3 lbs/ton

Pouring Temperature: 1575°C

(i) Differential Temperature Measurements:

<u>Casting</u>	<u>C%</u>	<u>Si%</u>	<u>Mn%</u>	<u>Remarks</u>	<u>Liquidus</u>	<u>Super-heat</u>
TG6A	0.27	0.12	0.27	Measurements along axis	1502°C	73°C
TG5E	0.28	0.10	0.36	ditto	1501°C	74°C
TG18A	0.27	-	-	Measurements across block	1502°C	73°C
TG19E	0.31	-	-	ditto	1498°C	77°C

(ii) Direct Temperature Measurements:

<u>Casting</u>	<u>C%</u>	<u>Si%</u>	<u>Mn%</u>	<u>S%</u>	<u>P%</u>	<u>Remarks</u>	<u>Liquidus</u>	<u>Super-heat</u>
A1(25A)	0.24	0.08	0.24	0.014	0.024	Measurements across block	1505°C	70°C
A2(38A)	0.29	0.10	0.43	-	-	ditto	1500°C	75°C
A3(33A)	0.24	0.20	0.36	-	-	Measurements along axis	1505°C	70°C
E1(28E)	0.28	0.09	0.34	0.018	0.018	Measurements across block	1501°C	74°C
E2(37E)	0.27	0.12	0.49	-	-	ditto	1502°C	73°C
E3(32E)	0.27	0.09	0.30	-	-	Measurements along axis	1502°C	73°C

The heads in all cases were removed and the block sectioned down the central vertical plane to determine accurately the positions of the thermocouples.

Examination of typical macrostructures showed that the introduction of the thermocouple sheaths had not affected the mode of crystallization. However, in some cases, generally when there had been inconsistencies in the cooling curves, it was found that slag had been trapped against one or more of the thermocouple sheaths, thus affecting the temperature readings. Such results were normally disregarded and the block recast.

(d) "Pour-out" Tests

In order to reproduce the conditions under which the temperature measurements had been determined, the same melting and casting procedures were employed for the "pour-out" experiments.

A series of four vertical top-fed blocks were cast; the "pour-outs" being effected after one, two, three and four minutes. The "pour-out" times represent the interval between the beginning of pouring and the final inversion of the mould. No correction was made for the time taken to invert the mould and for the liquid metal to drain out.

Finally the castings were sectioned parallel to the $2\frac{1}{2}$ " x 5" plane half way across the 3" face; this was the same plane as that in which the temperature determinations were made. The surface was machined, ground and etched in Oberhoffer's Reagent to reveal the macrostructure and the nature of the solidified metal. Scale drawings were constructed of the structures obtained and the depth of the solidified metal.

Examination of Castings

When cold, the moulds were broken open, the castings removed and the heads sawn off. The blocks were sectioned along the central vertical plane, in the as-cast state, and the surface ground for sulphur printing and/or macro etching. Depending on the results of the macroscopical examination, transverse sections were cut from the remaining portions of the blocks and similarly treated.

Samples for chemical analysis were taken from representative portions of the blocks to determine the average composition of the steel. The samples were not generally analysed for sulphur and phosphorus; as sillimanite pots were used no appreciable change in the contents of these elements would be expected.

Drawings were constructed showing the nature and extent of the different types of crystallization.

RESULTS.

(a) Size Effect.

Examination of the structure of the series of small sand castings (5" x 3" x 2½") showed that the structures were essentially similar to those of the series of larger commercially cast blocks (12" x 6" x 4"), described in an earlier section. This similarity indicates that the effects produced by varying the "positions" of the blocks and the mode of feeding, with consequential effect on convection currents are not restricted to castings of any particular dimensions. There were minor differences in the nature and extent of the different crystal zones, though the structures of corresponding blocks of the two series were as stated above, basically similar. Before considering the significance of such differences as were observed, the structures of the small castings will be described briefly and compared with the corresponding castings of the larger series.

Casting A (WOX6A, Fig. 52, 57).

The structure consists of a central nuclear zone surrounded by an envelope of columnar crystals. The columnar crystals extend from the side walls towards the centre for a depth of approximately 0.4 inches. Towards the top of the block the nuclear crystals coarsen, and small equiaxial dendritic crystals separate them from the columnar crystals.

A small pipe cavity extended from the head into

the upper portion of the block.

Comparing the structure of this block with those of the corresponding castings of the two commercial series (Figs.2 and 19), it will be seen to be roughly intermediate between that of A1 (which was completely nuclear) and A2 (which consisted of comparatively coarse columnar crystals enclosing a central zone of equiaxial crystals).

Casting B(WOX7, Fig.53,58).

The structure of this block consists of columnar crystals growing from all walls, surrounding a central region of small nuclear crystals. The columnar crystals have grown to a greater length from the top wall, than from the bottom. In the lower portion of the central area some of the nuclear crystals have developed into small dendritic equiaxial crystals.

Arising from the formation of a pipe there is a small area of unsoundness in the upper part of the block.

The structure of the small casting is very similar to that of one of the corresponding commercial castings (B2, Fig.22) but rather different from that of the other (B1, Fig.6), which consisted mainly of nuclear crystals with columnar crystals growing from the top face only.

Casting C(WOX3, Fig.54,59).

This casting consists of a central equiaxial and nuclear region surrounded by columnar growths which have developed from all the wall faces. That from the upper face

extends to the geometric centre of the block ($1\frac{1}{4}$ "), whereas from the lower face it only extends for about a third of the distance ($\frac{1}{2}$ "). In the central region the equiaxial zone lies directly below the upper columnar zone and is separated from the lower columnar growth by the nuclear zone. Both these zones extend from near the end wall into the head.

The pipe from the head extends for a small distance into the block along the junction between the upper columnar and the equiaxial zone. This corresponds to the last portion of the block to freeze.

A comparison between the structure of the larger castings C1 and C2 (Figs.8 and 25) and this smaller casting, shows that apart from more pronounced columnar growth from the lower face of the small casting, the structure of all three is very similar, there being only minor differences in the distribution of the central equiaxial and nuclear zones.

The structure of this small casting is also similar to those of a series of steel blocks, measuring 5" x 3" x 3", cast by West⁽¹⁶⁾ under similar conditions. There was a certain amount of unsoundness (Fig.59) in the present casting but in the light of the agreement with West's work it was considered unnecessary to recast the block.

Casting D (WOXB, WX2, Figs.55,60)

To avoid unsoundness it was found necessary to place exothermic sleeves around the head of this casting to

keep the head open and thus ensure adequate feeding during the solidification period.

Casting WOXB (Fig.55) was cast without the use of any exothermic compound. The structure consists of a central area of large equiaxial crystals surrounded by columnar crystals which have developed from the various wall faces. A central cavity extends from near the end wall, along the centre of the casting between the upper columnar growth and the equiaxial zone, into the head.

With the addition of exothermic cores a sound casting was obtained. The structure is essentially the same as that of the first casting, but without the central cavity. In the sound casting (WX2 Fig.55), the upper columnar crystals extend to the geometric centre near the end wall but become shorter near the head. The columnar crystals growing from the lower wall are also longer near the end wall than towards the head. Both the upper and lower columnar growths extend further towards the centre of the block in the sound casting than in the unsound one. The central area is wholly occupied by equiaxial crystals.

In the unsound casting WOXB (Fig.55), the pipe cavity assumes an unusual position in the vicinity of the junction between the feeder head and the base of the block. Instead of maintaining a central position roughly equidistant from the walls, it turns upwards towards the top face, finally reaching the mould wall. This seems to indicate that the sand

in the re-entrant angle of the casting (necessitated by the arrangement of the mould) acted as a local "hot spot", affecting the final heat centre and centre of freezing. The effect was less marked though not absent, in the sound casting WX2; it can be seen that towards the corner the depth of columnar crystals is considerably reduced, with a proportional thickening of the lower equiaxial zone. It is assumed that the junction between the upper columnar and equiaxial zones represents the position of the block last to freeze.

Comparing the structure of the small block with those of the larger castings D1 and D2 (Figs 11 and 28 respectively), it may be seen that although certain features are common to all three, there are notable differences in the distribution of the crystal zones. In all three castings there are columnar growths from the upper and end walls, but in the case of the smaller casting columnar crystals have also grown from the lower wall, and there is a marked relative increase in the length of the other columnar crystals. In the larger castings the central and lower portions of the blocks consist of zones of equiaxial and nuclear crystals, whereas in the smaller castings the central area is entirely equiaxial. The influence of the re-entrant angle at the junction between block and head, is reduced in the case of the larger castings, though in casting D1 the effect on the structure is still readily discernible.

Casting E (WOX5, WX1, WX3, (Figs 56, 61)

In order to obtain a sound casting, it was found necessary to incorporate exothermic cores at various positions around the head.

The first casting obtained (WOX5, Fig. 36), without the addition of exothermic cores, consisted entirely of columnar crystals, apart from a large central cavity. The columnar crystals extended towards the centre of the block, at right angles to the various wall faces. They were longer towards the head (at the base of the block) than at the top. At the bottom of the block there was a further area of unsoundness, concentrated on the side nearest the head.

The addition of exothermic sleeves around the head alone, did not eliminate the central cavity entirely. The structure of the block, (WX1 Fig. 56) was similar to that of casting WOX5. It consisted of columnar crystals, which had developed from the mould walls, together with a small area of nuclear crystals extending from the bottom of the block into the head. The area of the head situated immediately below the block consisted of a central region of nuclear crystals separated from a narrow zone of columnar crystals (growing from the walls of the feeder head) by a zone of mixed equiaxial and nuclear crystals.

The pipe cavity, like that of casting WOX8D (Fig. 55), assumes an unusual shape in the vicinity of the

junction between the feeder head and the base of the block. Within the block the cavity is equidistant from the walls, but towards the base it approaches the wall, finally breaking through to the surface, at the corner of the mould. It therefore seems that the sand in this region has acted as a local "hot spot", and has affected the position of the heat centre and that of the end of freezing.

The addition of an exothermic core, in the form of a plate, to the bottom of the head, successfully eliminated the central cavity. The structure of the block (WX3) is similar to that of the earlier castings. It consists entirely of columnar crystals apart from a small area of equiaxial crystals at the base of the block. The columnar crystals extend at right angles to the wall towards the centre, meeting along the central axis.

The addition of the exothermic core along the base of the feeder head has naturally affected the heat distribution in this region, and also the resulting crystal formation. The structure consists of columnar crystals growing from the walls, with a pile of nuclear crystals resting on them and developing into small equiaxial crystals which extend into the block.

By lowering the position of the heat centre and the final position of freezing, the introduction of the exothermic plate has also affected the position of the pipe cavity. It is now confined to the head and is located between

the lower nuclear zone and the upper columnar growth.

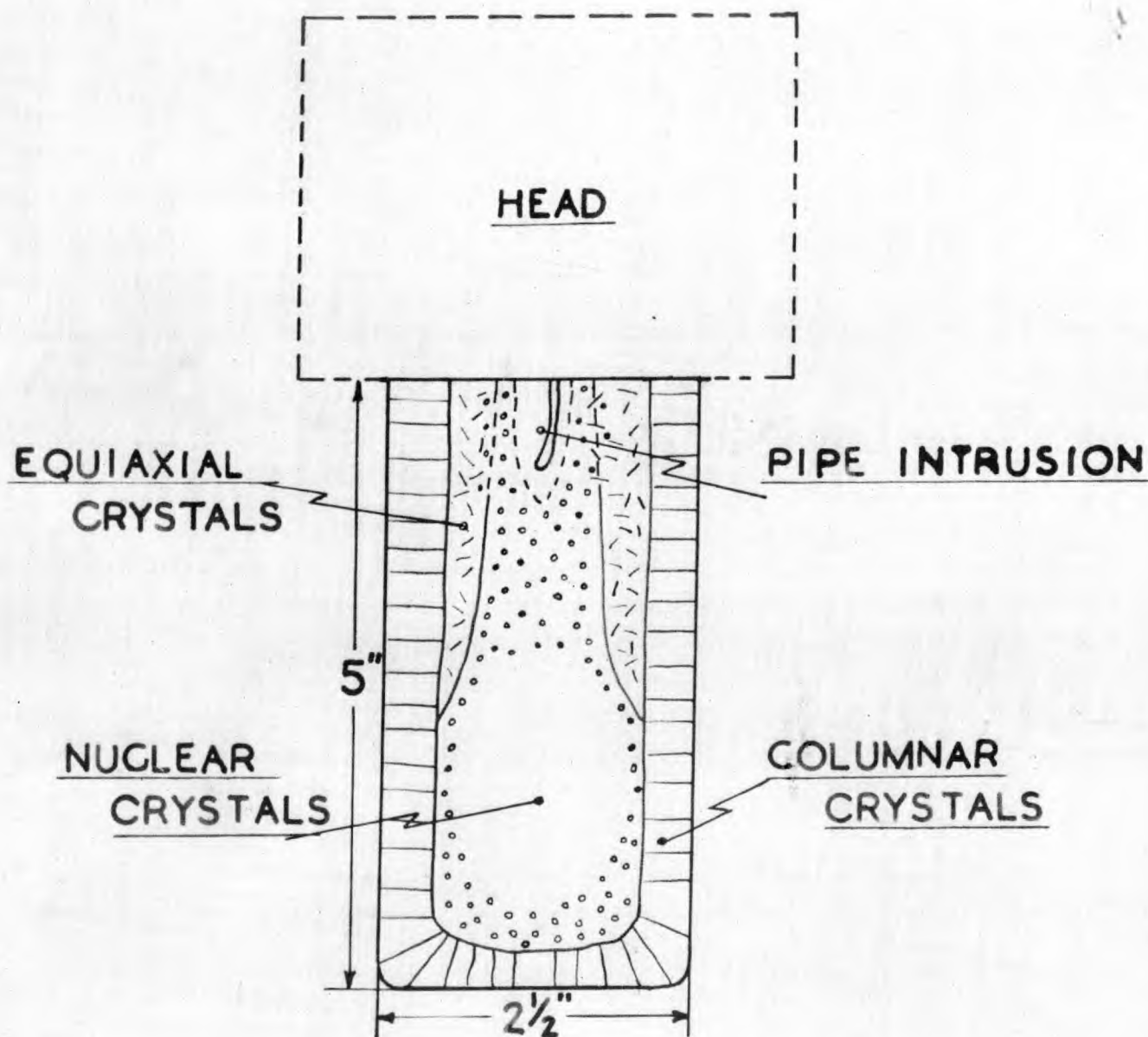
If the structure of the small casting WX3 is compared with that of the larger casting E1 (Fig.15), it will be seen that they are basically similar, but that the central zone of equiaxial crystals, which appears only at the bottom of the small block, extends almost to the top of the large one. These structures cannot be compared with the anomalous structure of casting E2, (Fig.31).

Transverse Sections.

During the examination of this series of small blocks, transverse sections were prepared and examined. The structures obtained were in accordance with the corresponding longitudinal sections and with the transverse sections of the larger series described in Section 2, the only feature requiring comment being that the structures indicated that solidification from the side walls had not affected the crystallization at the centre.

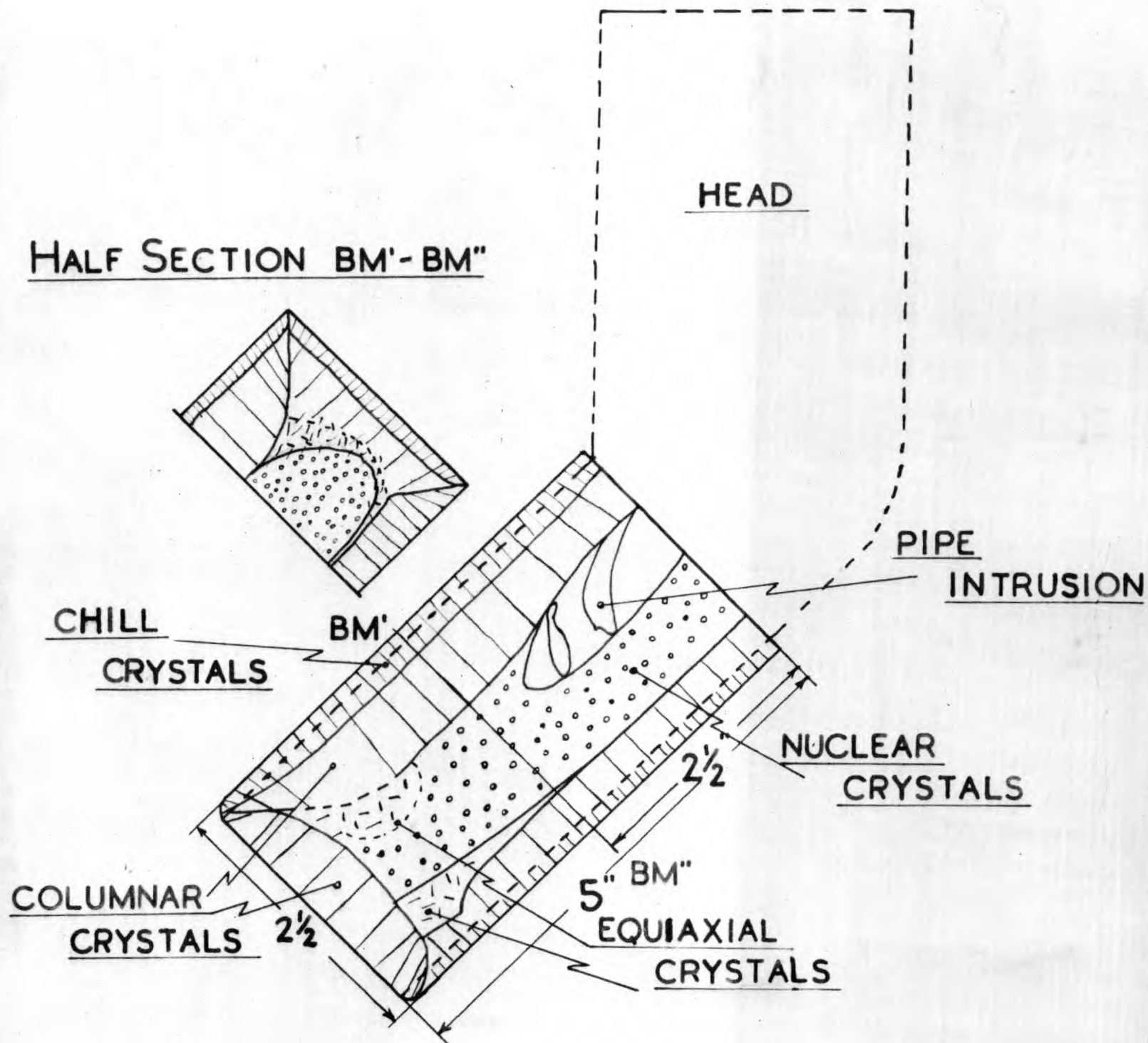
Preliminary Comments on Results.

If the structure of the series of small castings is considered in conjunction with those of the larger castings of series I, it will be found that a complete range of structures has been obtained from the wholly nuclear structure of A1 (Fig.2) to the almost wholly columnar structure of WX3 (Fig.56). The significance of this finding will be discussed in Section 4 of the thesis after the additional data provided by the other series of experiments has been presented.



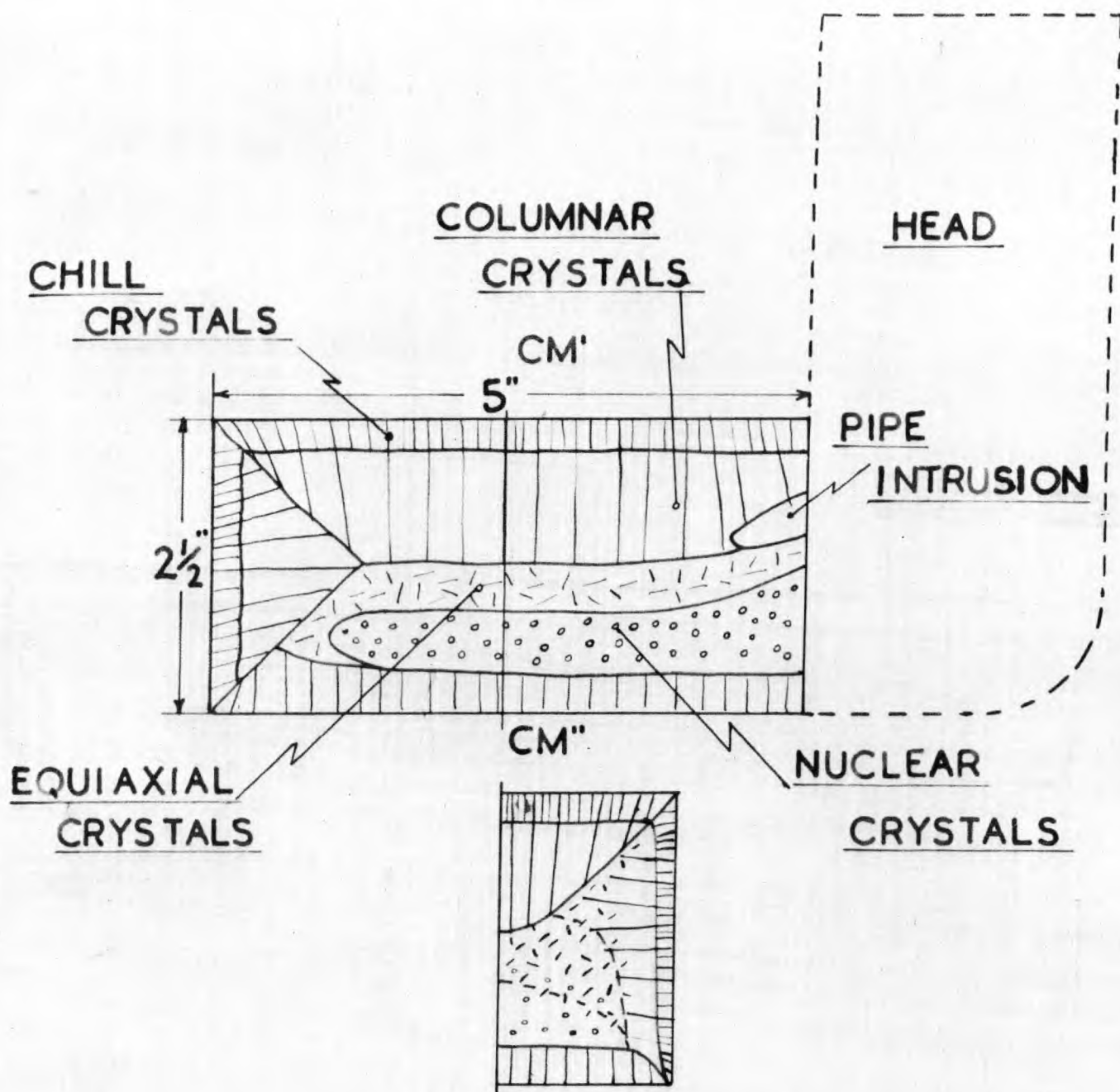
STRUCTURE OF CASTING 'A'

FIG.52



STRUCTURE OF CASTING 'B' WOX7

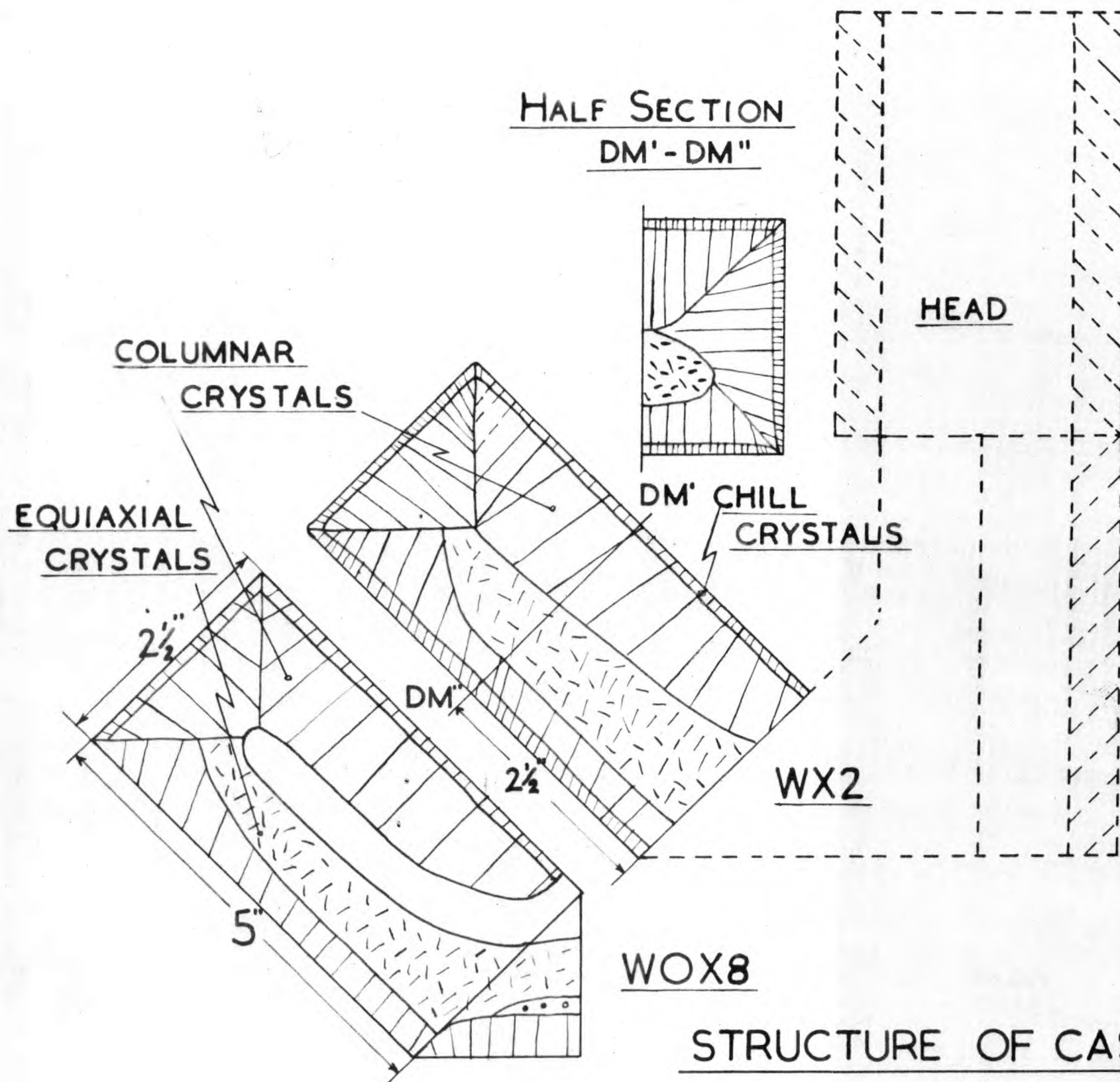
FIG. 53



HALF SECTION CM'-CM"

STRUCTURE OF CASTING 'C' WOX3

FIG. 54

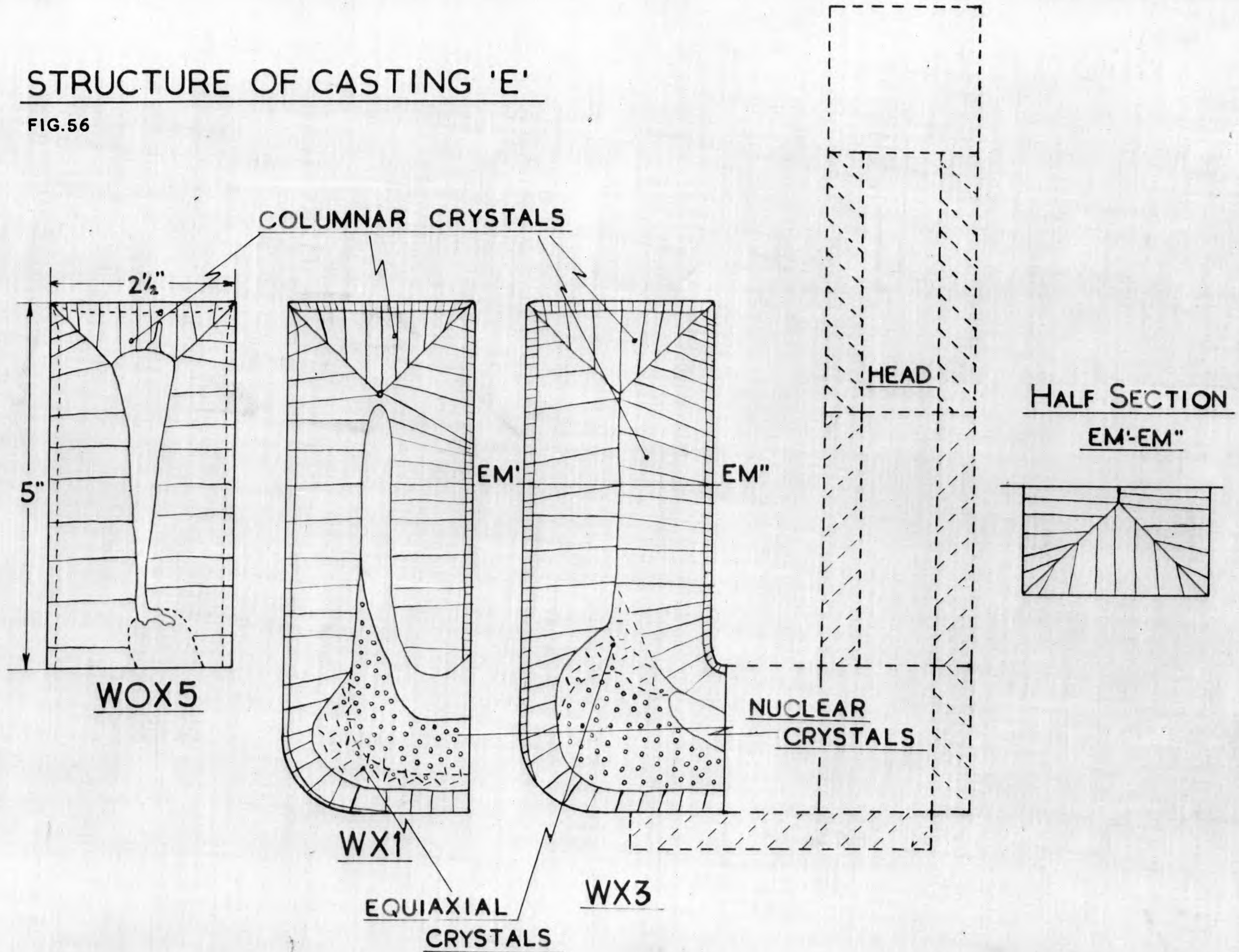


STRUCTURE OF CASTING 'D'

FIG. 55

STRUCTURE OF CASTING 'E'

FIG.56



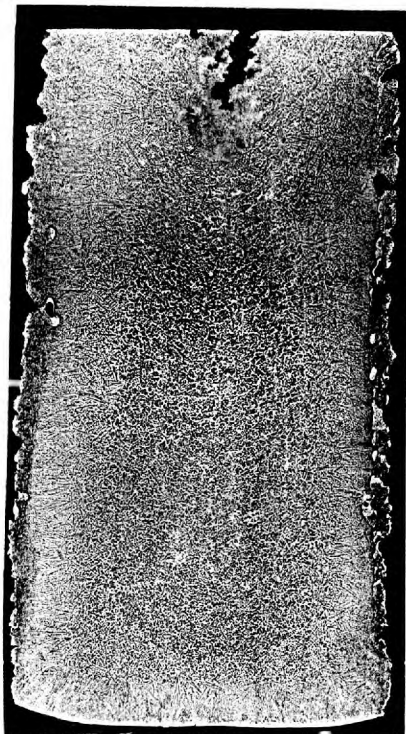


Fig. 57. Macrostructure of
Casting WOX6A.

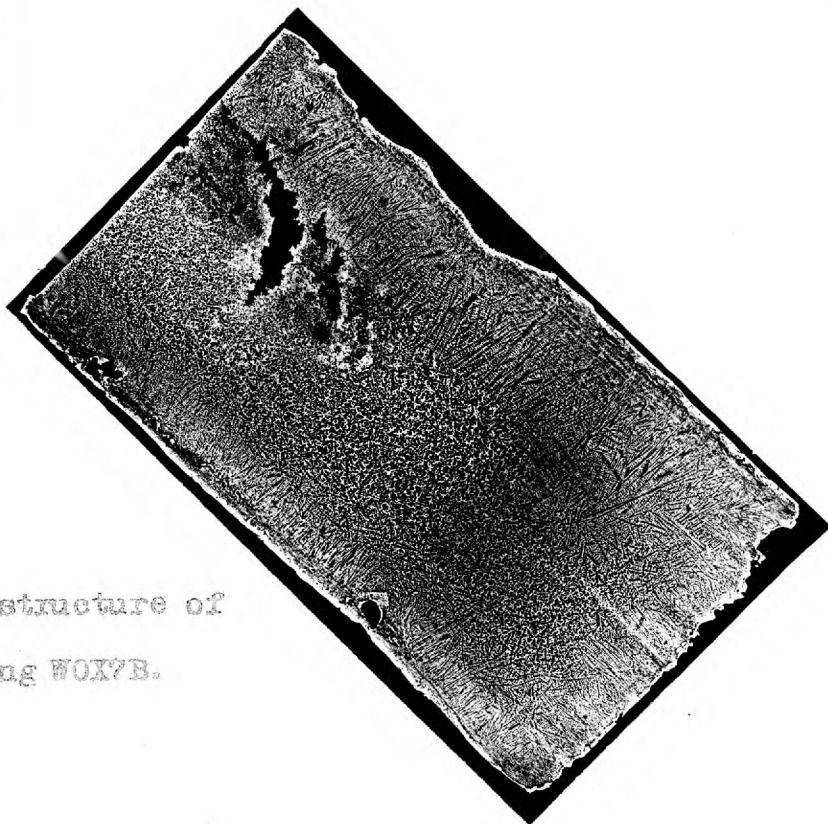


Fig. 58. Macrostructure of
Casting WOX7B.



Fig. 59.
Macrostructure of
Casting WOI3C.

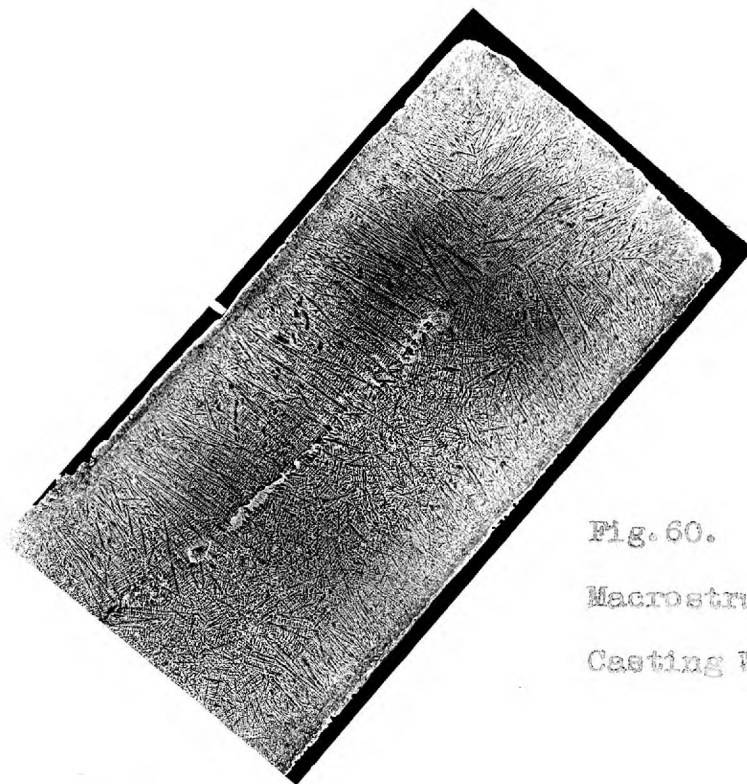


Fig. 60.

Macrostructure of
Casting WX2D.

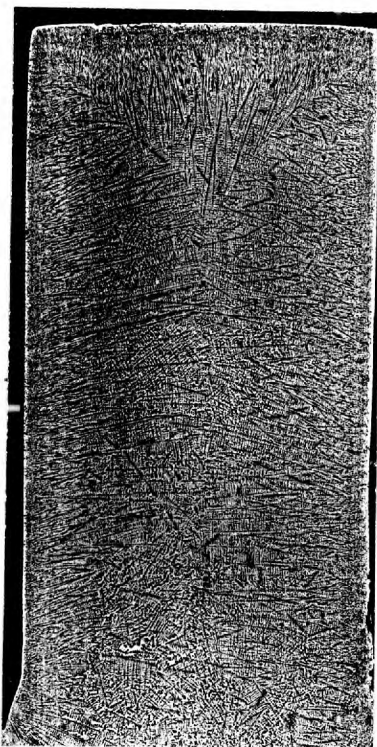


Fig. 61. Macrostructure of
Casting WX3E.

(b) Casting Temperature.

An investigation of the effect of casting temperature has shown that this variable has a marked effect on the crystallization of sand castings. Six experimental top-fed vertical castings were made, with superheats ranging from 0 to 125°C above the liquidus temperature. The structures obtained varied from a completely nuclear structure to one consisting of a central mass of equiaxial dendritic crystals surrounded by a thick columnar zone. The individual results are described below.

Casting SHA1. Casting Temperature 1500°C (Fig.62,67).

Calculated Superheat 0°C. *

The entire structure of this casting consists of fine nuclear crystals. Towards the top there are three areas of unsoundness which have resulted from slag inclusions and pre-solidified material.

Casting SHA2. Casting Temperature 1525°C (Fig.63,68).

Calculated Superheat 25°C.

The structure of this casting consists of a thin shell of small columnar crystals (0.2" thick) surrounding a region of fine nuclear crystals. Towards the top of the block there is a small area of unsoundness due to the pipe cavity.

* Details of the chemical composition of the block and the "actual" superheat value are given on page 90.

Casting SHA3. Casting Temperature 1550°C. (Figs. 64, 69).

Calculated Superheat 50°C.

The structure of this casting consists of a shell of columnar crystals (0.3" thick) and a central region of nuclear crystals. Between these two zones is a zone of small equiaxial dendritic crystals, which extend from the top of the block towards the bottom; the zone is thicker at the top than at the bottom. There is also a small area of unsoundness, due to the pipe, at the top of the block.

Casting SHA4. (WOX6A). Casting Temperature 1575°C.

Calculated Superheat 75°C. (Figs. 52, 57).

A block was cast under the above conditions during the investigation of the effect of size factor (Page 97). The block, as previously described, consists of a shell of columnar crystals enclosing a region of mainly nuclear crystals. Towards the top of the block and extending down the side of the columnar crystals are two areas of small equiaxial crystals. Although they do not appear to extend as far down the block as in casting SHA3 (Fig. 64), the equiaxial areas appear to be slightly thicker at the top and the columnar shell is also thicker in all positions.

There is again a small area of unsoundness, due to the pipe from the head, at the top of the block.

Casting SHA5. Casting Temperature 1600°C.

Calculated Superheat 100°C. (Figs.65,70).

The structure of this block consists of a shell of columnar crystal approximately 0.5" thick, enclosing a central mass of nuclear and equiaxial crystals. The equiaxial zone is situated toward the bottom of the block, with the nuclear zone above it.

Casting SHA6. Casting Temperature 1625°C.

Calculated Superheat 125°C. (Figs.66,71).

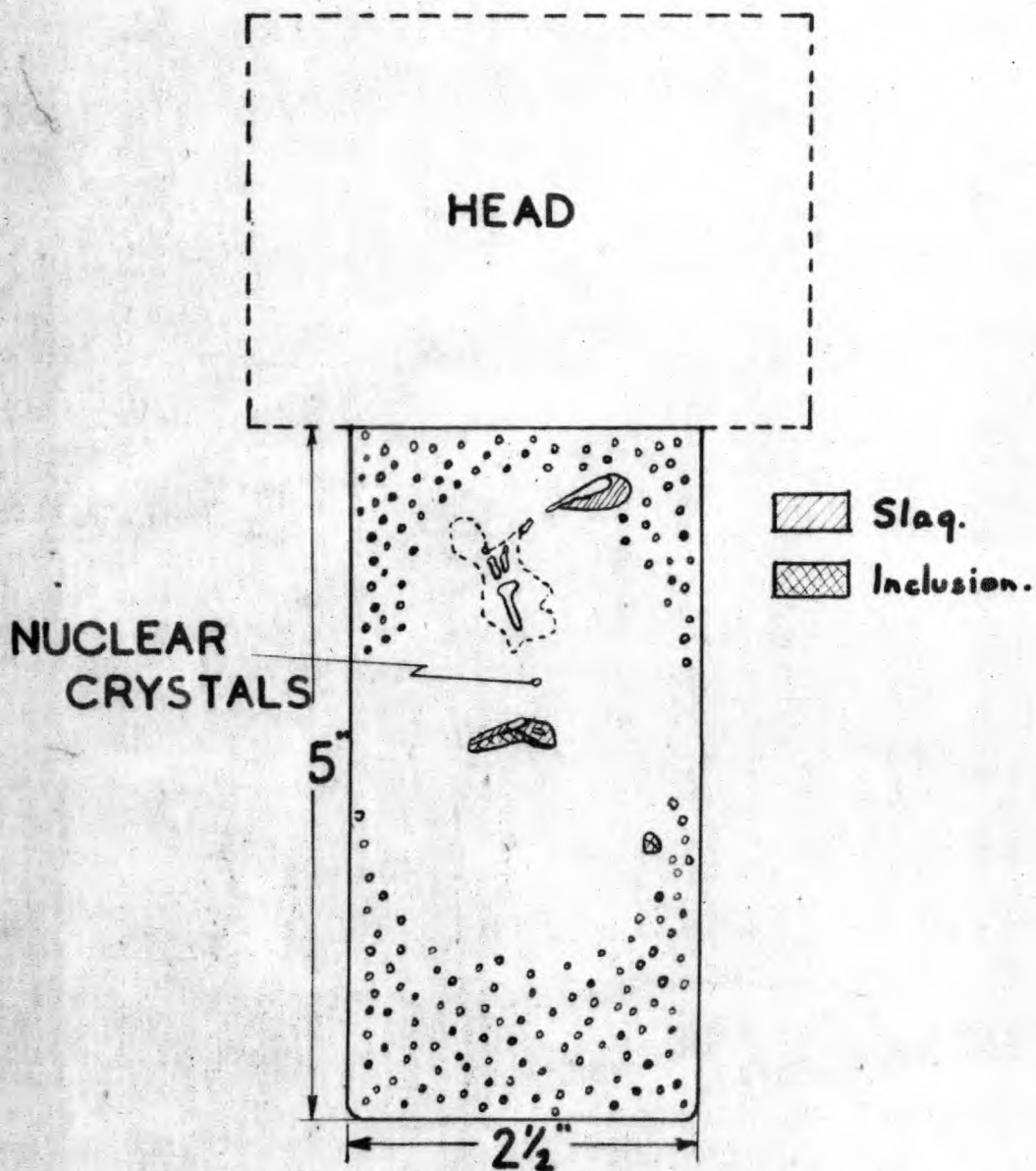
The structure of the final casting consists of a thick shell of columnar crystals, approximately 0.6" thick, enclosing a central area of nuclear and equiaxial crystals. The equiaxial zone occupies the major portion of the central area, and again the nuclear zone is above it.

Preliminary Comments on the Results.

A comparison of the structures of the series of castings shows that the increase of pouring temperature has had two main effects on the form of crystallization. Firstly, it has resulted in the development of columnar crystals from all the walls. As the temperature has been raised the length of the columnar crystals has increased from 0" when cast at the liquidus temperature, to approximately 0.6" when cast at a superheat of 125°C. Secondly, it has resulted in a steady coarsening of the structure. The first casting, poured at the liquidus temperature, consists of fine nuclear crystals, whereas the structure of the central portion of the block

cast at 125°C superheat consists largely of coarse equiaxial dendritic crystals.

The intermediate castings show a transition between the two extreme forms.



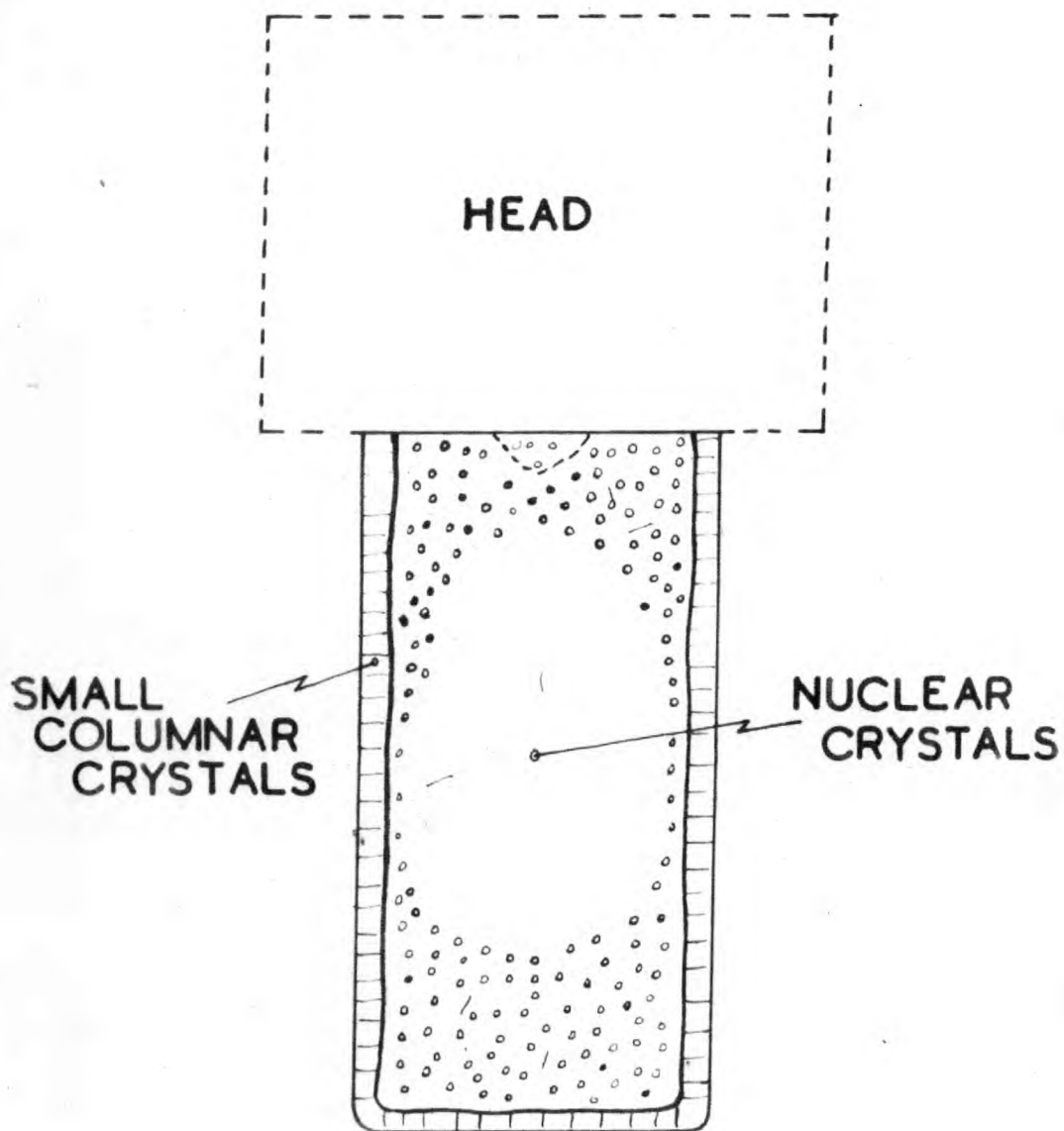
SHAI STRUCTURE OF CASTING A

CASTING TEMP. 1500°C

LIQUIDUS TEMP. 1500°C

FIG.62

N/S.



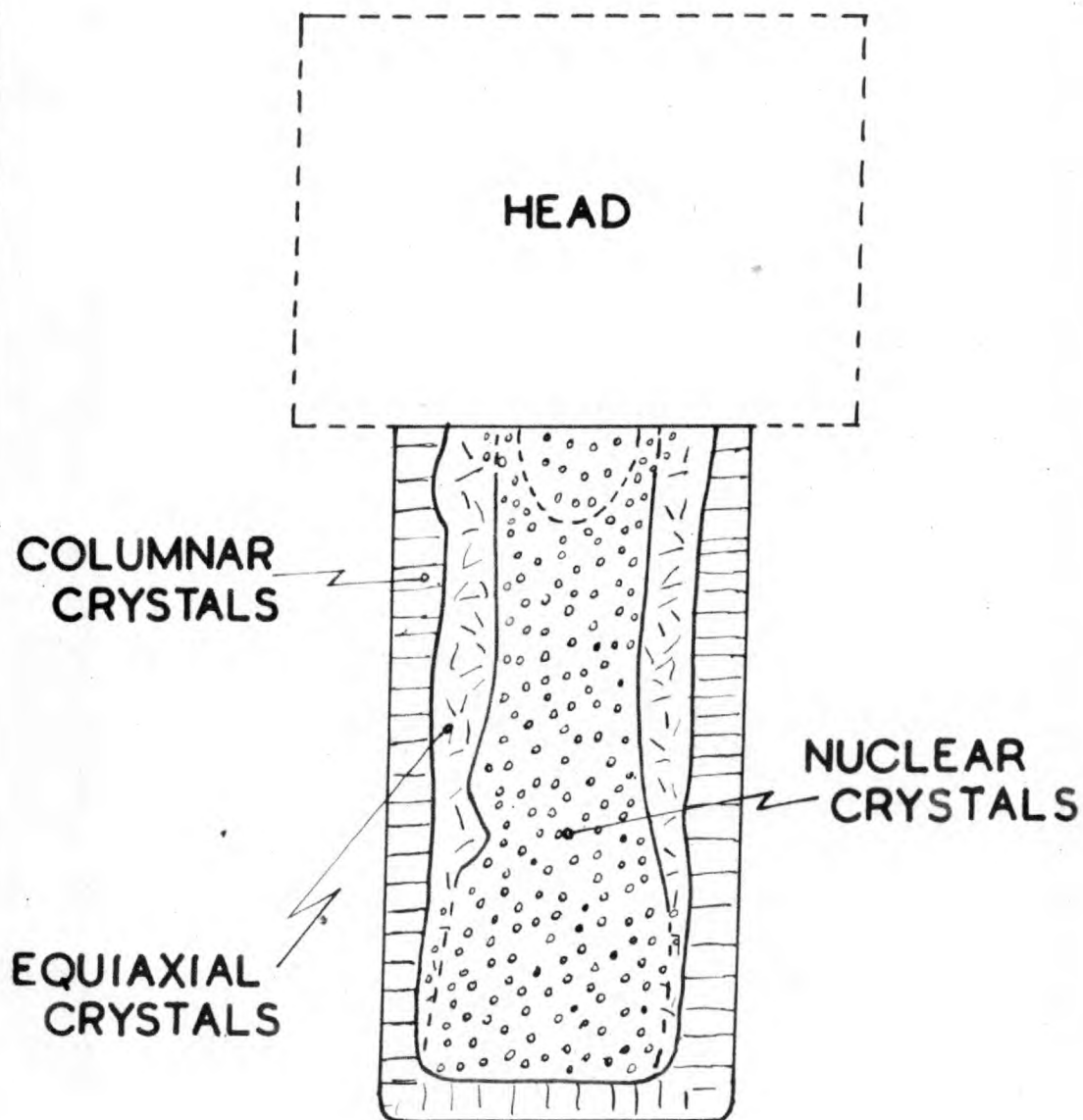
SHA2

STRUCTURE OF CASTING A

CASTING TEMP. 1525°C

LIQUIDUS TEMP 1500°C

FIG.63

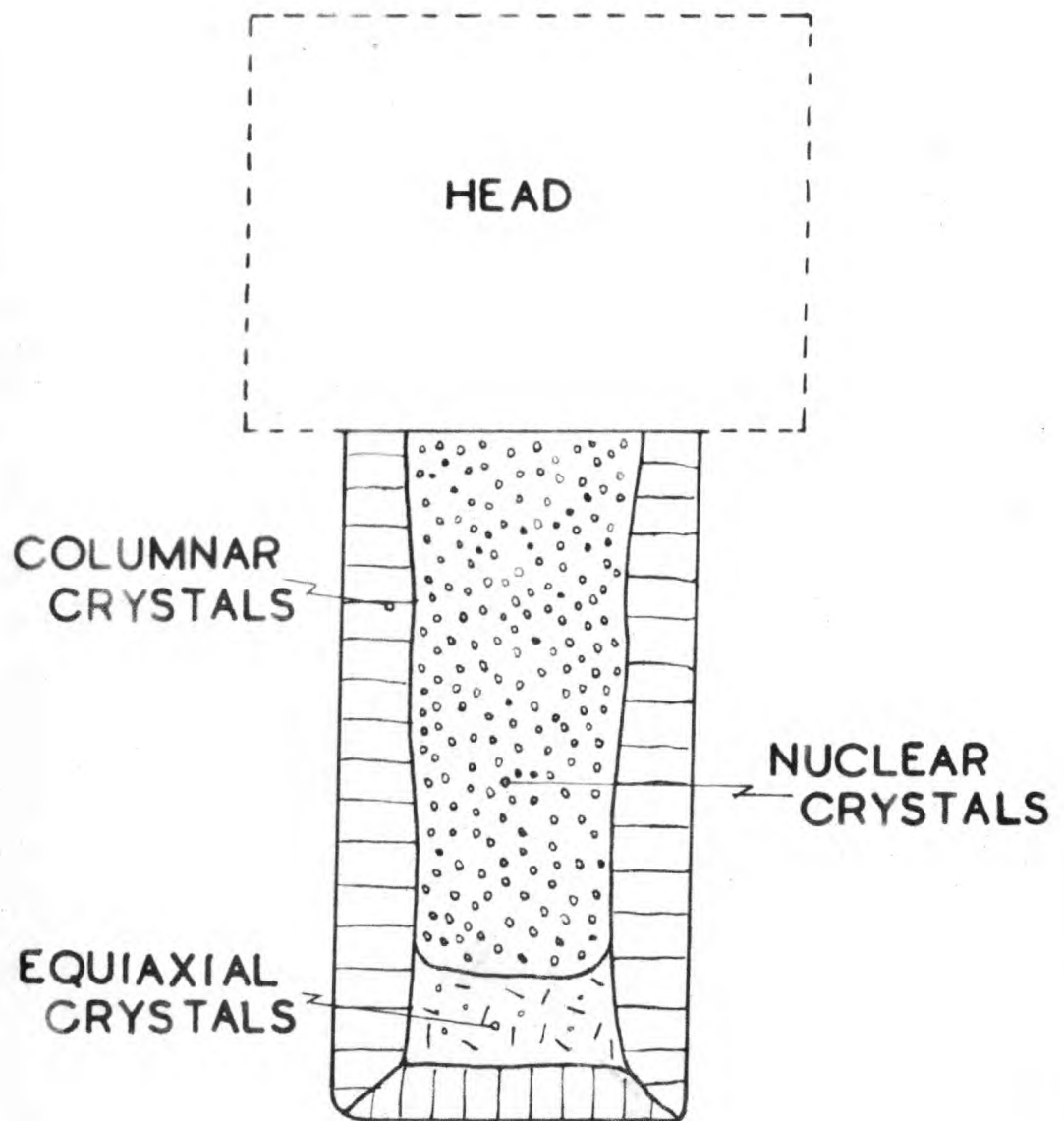


STRUCTURE OF CASTING A

CASTING TEMP. 1550°C

LIQUIDUS TEMP. 1500°C

FIG.64

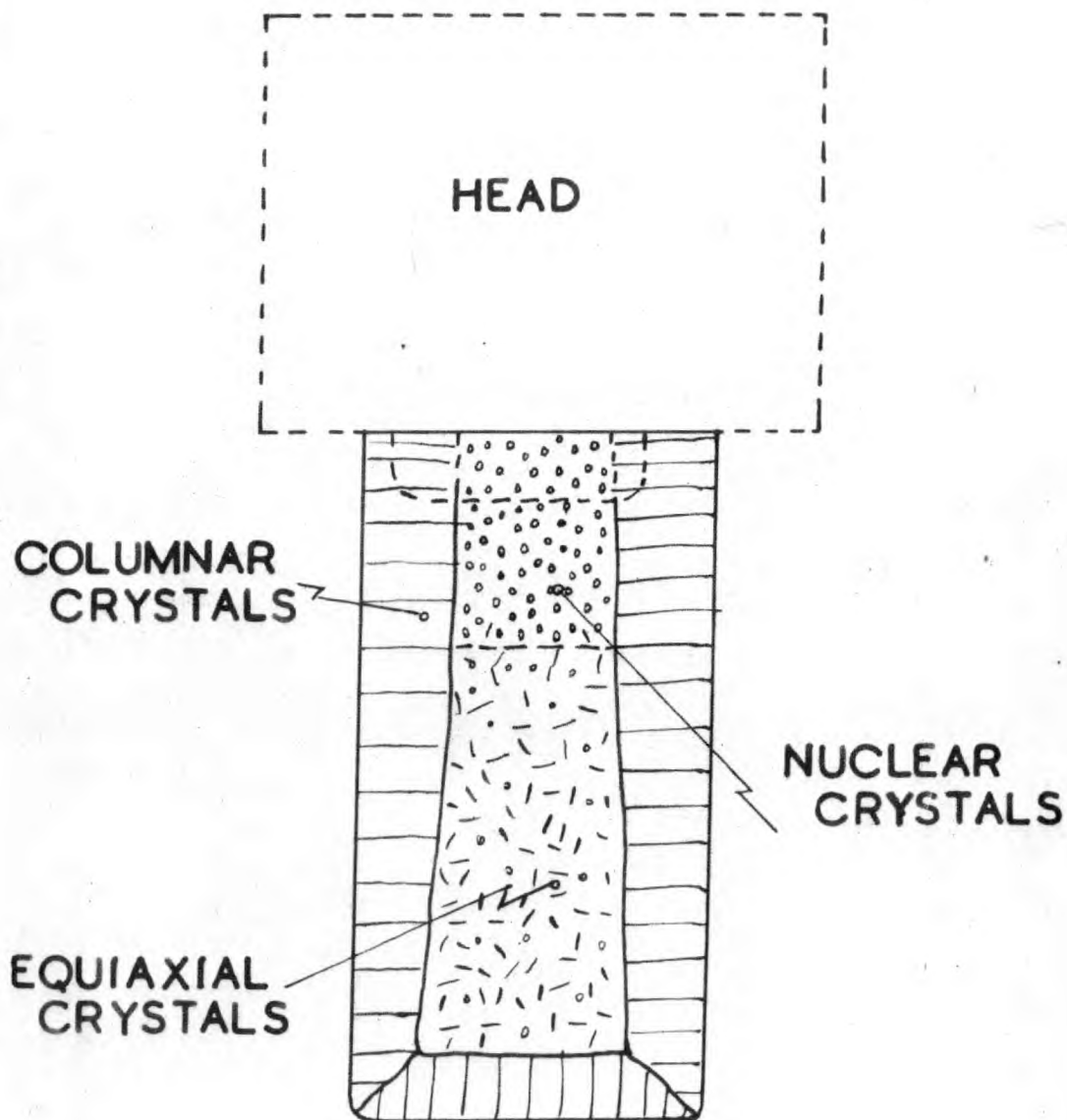


SHA5
STRUCTURE OF CASTING A

CASTING TEMP. 1600°C

LIQUIDUS TEMP. 1500°C

FIG. 65



SHA6

STRUCTURE OF CASTING A

CASTING TEMP. 1625°C

LIQUIDUS TEMP. 1500°C

FIG.66



Fig.67. Macrostructure
Casting SHA1.

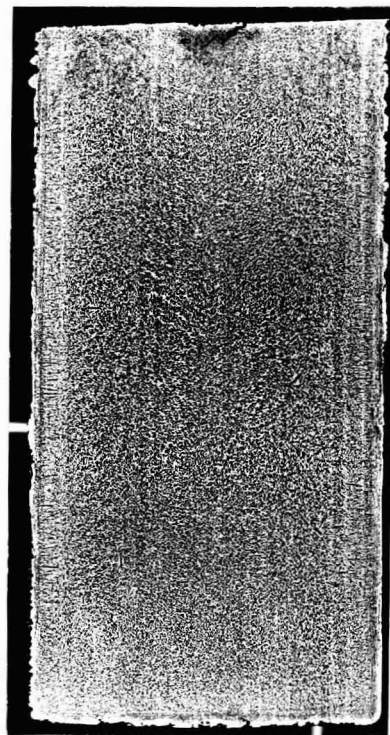


Fig.68. Macrostructure
Casting SHA2.

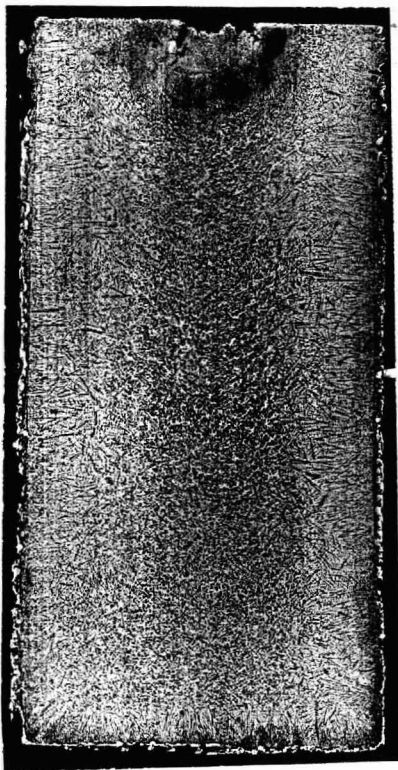


Fig.69. Macrostructure
Casting SHA3.

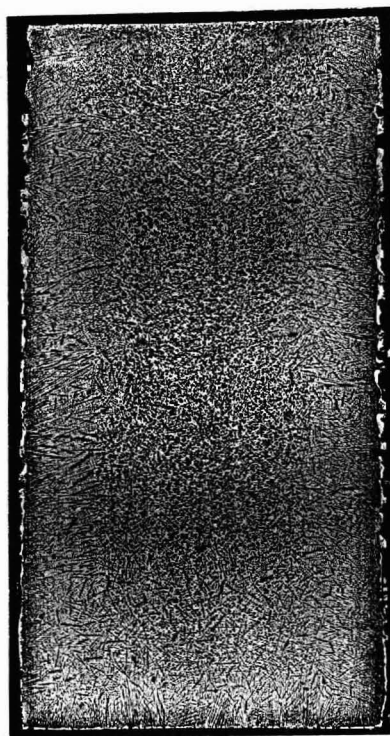


Fig.70. Macrostructure
Casting SHA5.

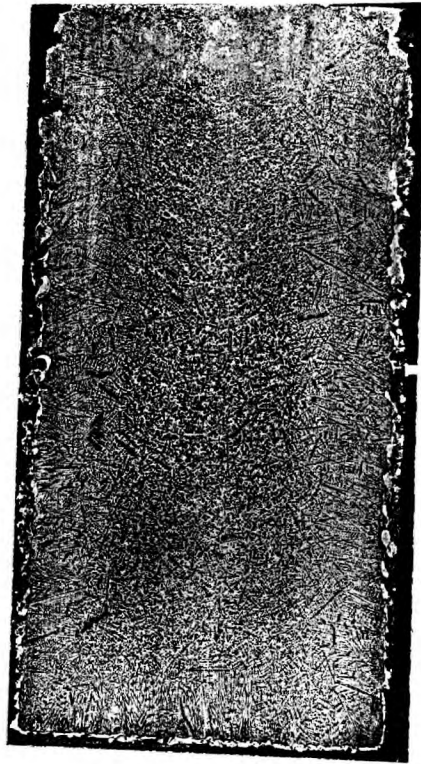


Fig.71. Macrostructure
Casting SHA6.

(c) Temperature Measurements. (1) Differential Method.

Before the data provided by this method could be used it was necessary to translate the information obtained from the recording apparatus into units of temperature and time. This was achieved by constructing two scales - one representing temperature difference (in degrees Celsius) in terms of the deflection of the recorder needle, and the other representing time in terms of the distance travelled by the recorder chart. Briefly, the scales were constructed as follows. Certain standard e.m.f.s. were applied to the recorder, and the corresponding deflections of the recorder needle (which depended, of course, on the range being used) were noted. With this information, and by reference to standard calibration tables for the platinum - rhodium thermocouples, it was possible to construct a scale representing temperature differences in terms of e.m.f. and hence deflection of the needle, at a given temperature. As the temperature/e.m.f. relationship is not constant over a range of temperature, two scales were prepared, one for use in the region of 1500°C and another in the region of 1400°C . The difference between the two scales was about 2%.

The scale relating time in terms of distance travelled by the recorder chart was constructed by measuring the distance travelled in a given time. (One minute corresponded to $7\frac{1}{2}''$ travel).

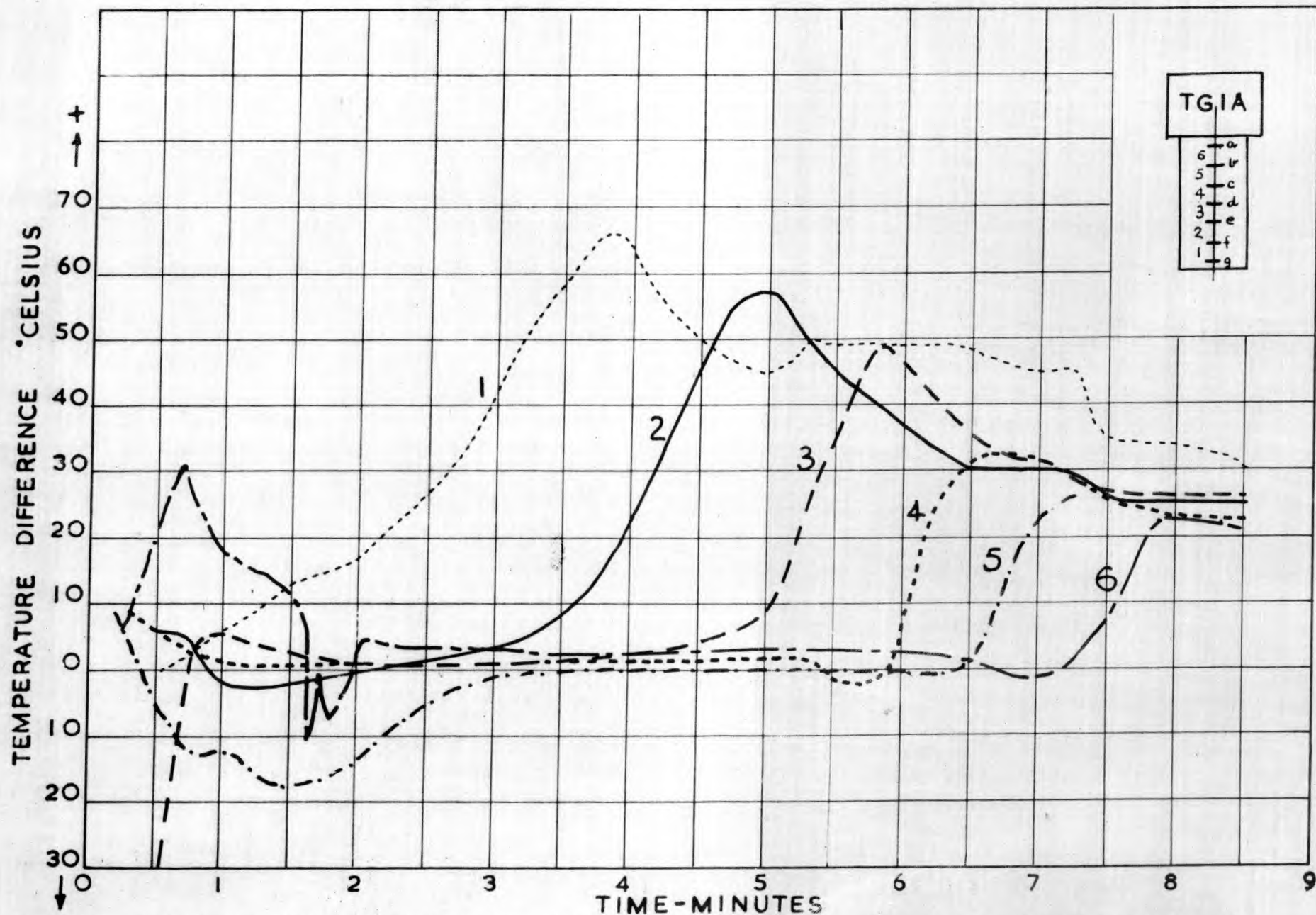


FIG.72 TEMPERATURE DIFFERENCE CURVES OBTAINED BETWEEN POSITIONS ON CENTRAL AXIS OF TOP FED VERTICAL BLOCK.

By using these scales, curves representing the temperature gradients existing during the solidification period were prepared. These are shown in Figs. 72 - 75.

The curves of temperature difference obtained for the vertical castings TGLA and TGLE (Figs. 72, 73), represent the temperature differences along the central axis of the top and bottom-fed blocks respectively. Those for castings TG2A and TG2E (Figs. 74, 75), represent the temperature differences existing horizontally across the top-fed and bottom-fed blocks respectively.

Although it has been found that these data cannot be used for determining the "absolute" temperature conditions existing during solidification, due to additive errors, several interesting features are, nevertheless, revealed by examining the different sets of curves.

Casting TGLA (Fig. 72).

The curves obtained from this casting suggest that there are three stages in the progress of solidification along the axis of the block.

The first stage lasts for about 1 - 2 minutes, during which time erratic variations in the temperature are registered by the various differential couples (1, 2, 3 etc.). Ultimately, however, the erratic fluctuations are smoothed out, the temperature differences tending to zero.

During the second stage, a series of maxima

occur, in one curve after the other corresponding to the progression of solidification along the length of the block, from the bottom towards the head. Curve 1 shows that there is a steady increase in the temperature differences between positions "g" and "f" near the bottom of the block, reaching a maximum value of 66°C in 3 minutes 55 seconds; after this it decreases somewhat irregularly. After about 2 minutes, curve 2 shows an accelerating increase in temperature difference between positions "f" and "e", reaching a maximum value of 57°C in 5 minutes, after which it steadily decreases. Similarly, curves 3, 4, 5 and 6 show successive increases and decreases in the temperature differences between positions "e" and "d", "d" and "c", "c" and "b" and finally "b" and "a" near the head. In each case the maximum temperature difference recorded is less than the preceeding value. During this period there appears to be little or no temperature gradient from the top of the block to the various positions down the axis until the curve for each successive position begins to rise steeply, presumably owing to the start of solidification. Immediately before the steep rise in curves 4, 5 and 6, after 6, $6\frac{1}{2}$ and 7 minutes respectively, there is a distinct dip in each curve indicating a temporary temperature difference in the reverse direction.

The third stage in the solidification process begins when the temperature difference between the top two

positions on the axis of the block has reached a maximum, and continues until the block has cooled to room temperature. During this period the temperature differences between the various positions along the axis of the block are approximately equal, decreasing slowly and steadily as cooling proceeds.

When the direct cooling curve obtained from the thermocouple situated nearest the head (curve not shown) was compared with the sets of difference curves, it was found that the maximum temperature difference between positions a and b (curve 6, Fig.105), occurred approximately at the peritectic temperature for that position. The significance of this is discussed later on Page 162. No feature of the temperature-difference curves seems to correspond with the time of the end of the liquidus arrest on the direct curve.

Pos ⁿ	Distance of couples from End Wall.	Beginning of Rise in Temp. Diff. Curve.	Max. Temp. Differences.	
			Time.	Temp.
1	Between g (0.40") and f (1.08")	15 seconds	3m. 51 secs.	66°C
2	Between f (1.08") and e (1.82")	2 minutes	5m. 0 secs.	57°C
3	Between e (1.82") and d (2.62")	3 min. 45 secs.	5m. 51 secs.	48°C
4	Between d (2.62") and c (3.35")	5 min. 57 secs.	6m. 45 secs.	32°C
5	Between c (3.35") and b (4.08")	6 min. 27 secs.	7m. 24 secs. (approx.)	26°C
6	Between b (4.08") and a (4.80")	7 min. 21 secs.	7m. 57 secs.	23°C

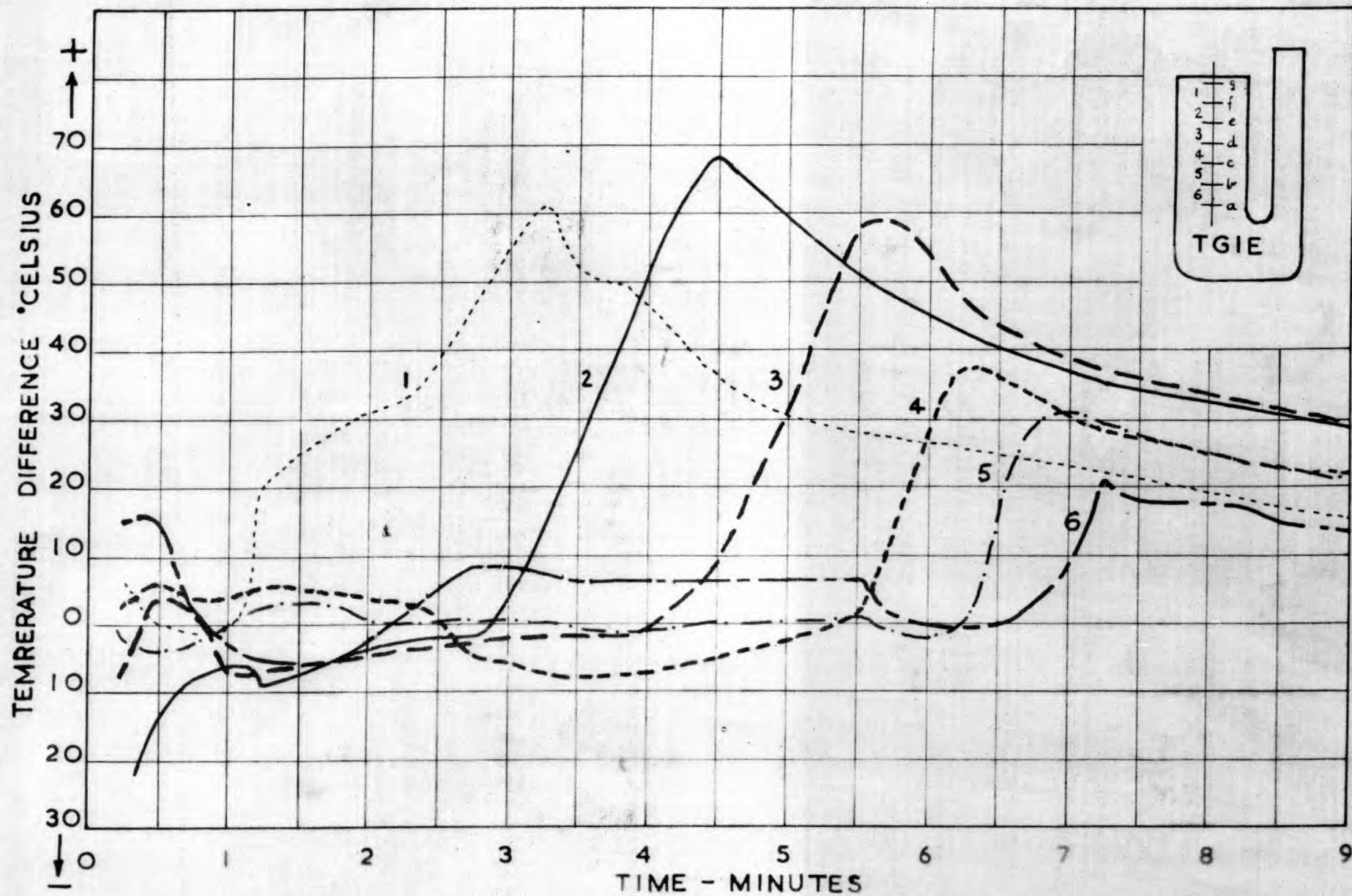


FIG.73 TEMPERATURE DIFFERENCE CURVES OBTAINED BETWEEN POSITIONS ON
CENTRAL AXIS OF BOTTOM FED VERTICAL BLOCK

Casting TGLE (Fig.73.)

The corresponding set of temperature difference curves obtained during the solidification of the bottom-fed casting TGLE shows similar features to those described for the top-fed casting TGLA, for example the same three stages during the solidification process. There are, however, some points of difference.

The slope and position of the maxima on the temperature-difference curves are different. Although the increase in the temperature difference between positions g and f (on the axis of the block furthest away from the head) does not begin until 1 minute after pouring, the rise during the next 15 seconds is very rapid; after this the increase in temperature difference is slower, reaching a maximum of 60°C after 3 minutes 15 seconds. The temperature difference between f and e increases after about 3 minutes, rising to a maximum of 68°C after 4 minutes 30 seconds. This maximum is higher than that of curve 1 and is 10°C higher than for the corresponding position in casting TGLA. A similar series of peaks is shown by curves 3,4,5 and 6, corresponding to the other positions along the axis of the block. They are individually higher than the corresponding series of peaks for casting TGLA. Although in comparison with TGLA, the rise of curve 1 (furthest from the head) begins approximately 30 seconds later, the maximum on curve 6 (nearest the head) is

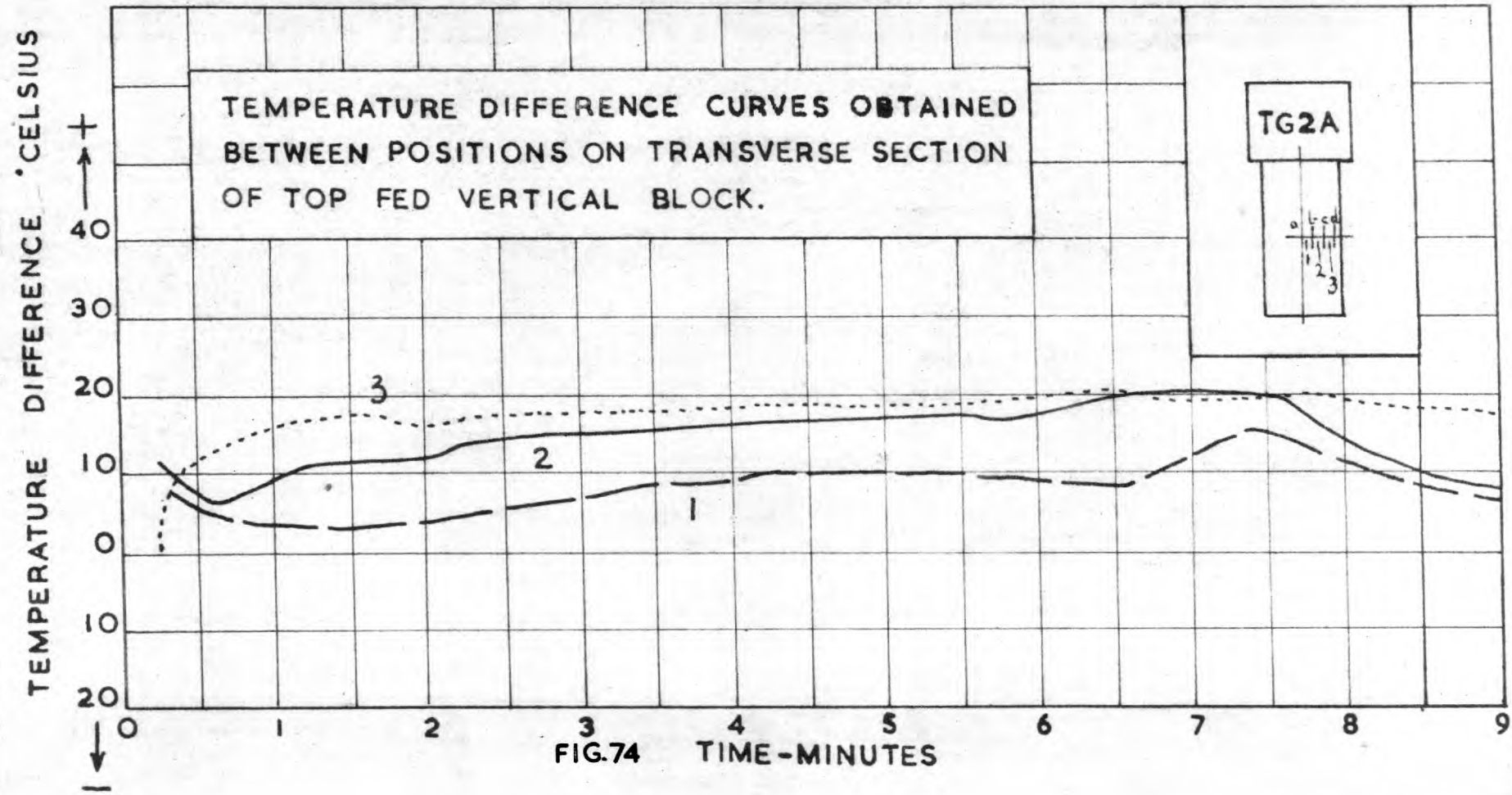


FIG.74

TIME-MINUTES

attained over 30 seconds earlier. This indicates that casting TGLE solidified faster than casting TGLA.

The "reversals" in temperature difference observed in curves 4, 5, and 6 of casting TGLA, immediately before the rise towards the maximum values, are repeated in curves 5 and 6 of the present series.

After the final increase in temperature difference, as shown by curve 6, cooling proceeds steadily to room temperature. During this period the temperature gradients are steeper near the end of the block than near the head, but they decrease as cooling proceeds.

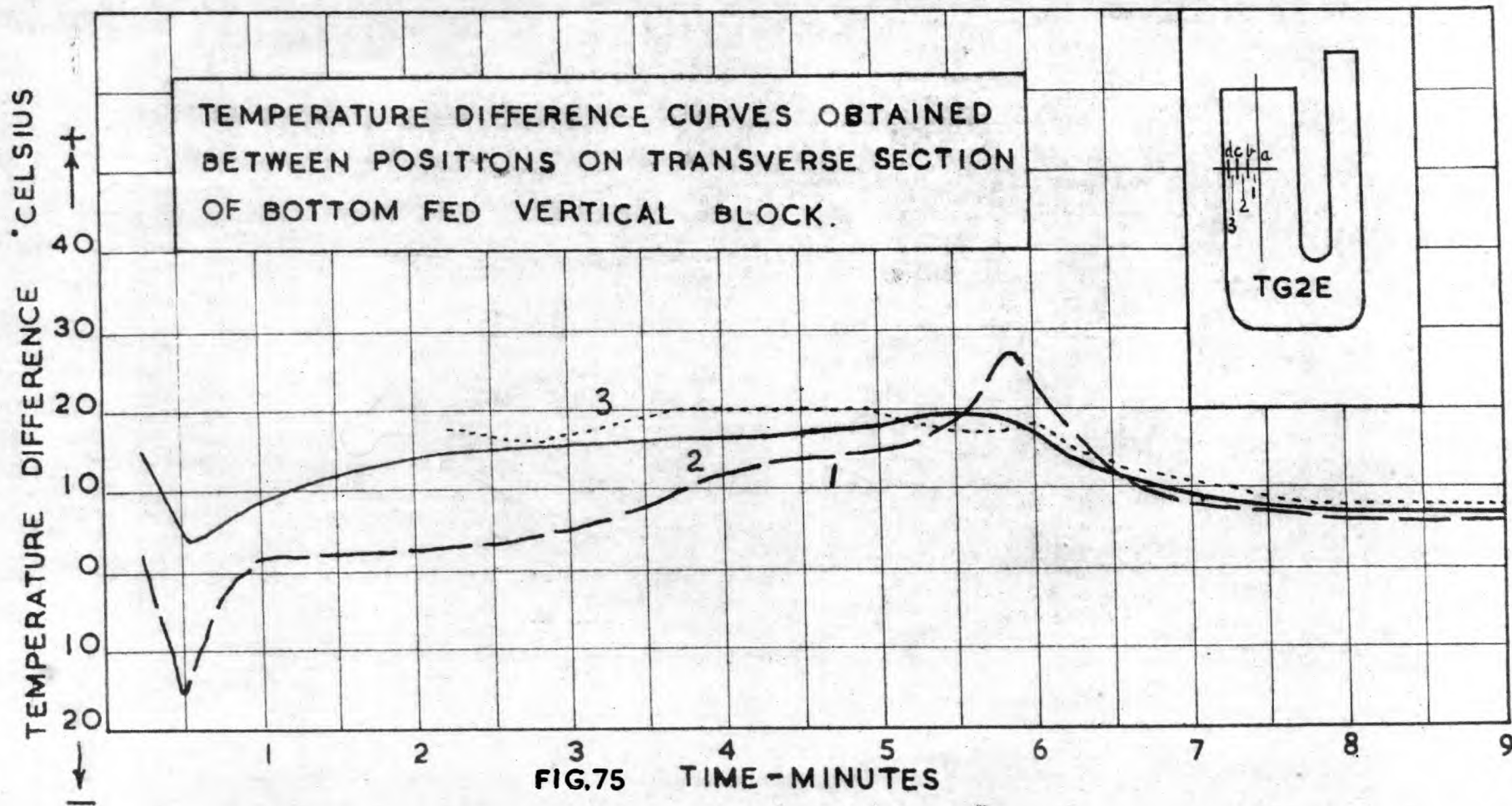
Pos ⁿ .	Distance of couples from End Wall.	Beginning of Rise in Temp. Diff. Curve.	Max. Temp. Differences.	
			Time.	Temp.
1	Between g (0.23") and f (0.80")	1 min. 3 secs.	3m.18 secs.	61°C.
2	Between f (0.80") and e (1.48")	2 min.50 secs.	4m.30 secs.	68°C.
3	Between e (1.48") and d (2.26")	4 min. 0 secs.	5m.39 secs.	59°C.
4	Between d (2.26") and c (3.00")	5 min.30 secs.	6m.18 secs.	37°C.
5	Between c (3.00") and b (3.70")	6 min.15 secs.	7m. 0 secs.	31°C.
6	Between b (3.70") and a (4.42")	6 min.45 secs. (approx.)	7m.15 secs.	21°C.

Casting TG2A (Fig.74).

The curves obtained by measuring the temperature difference across the middle section of a top-fed block during solidification are quite unlike those obtained along the central axis (Casting TGLA). This may be ascribed, partly at any rate,

to the fact that in measuring the temperature differences in the transverse plane, the thermocouples were placed only about $\frac{1}{4}$ " apart, whereas in measurements along the axis they were at $\frac{3}{4}$ " intervals.

The curves show that after an initial period of about 15 seconds to about $1\frac{1}{2}$ minutes, depending on the position in the block, the temperature differences between the various positions increase from a relatively low value to a maximum of $17^{\circ} - 20^{\circ}$ C., and then decrease as cooling proceeds. The time taken for the maximum difference to be attained is related to the position of the thermocouples within the solidifying block. The temperature difference between positions c and d situated near the mould wall, increase rapidly to 17° C after $1\frac{1}{2}$ minutes from pouring. After $7\frac{1}{2}$ minutes it has only increased a further 3° C to a total difference of 20° C. The temperature difference between the two central positions, a and b, increase in two stages. After increasing gradually to 10° C, $4\frac{1}{2}$ minutes after pouring, it then remains constant until about $6\frac{1}{2}$ minutes, after which time it rises rapidly to a maximum of 16° C at $7\frac{1}{2}$ minutes. Curve 2 representing the temperature difference between the intermediate position b and e, rises gradually to a value of 15° C in $2\frac{1}{2}$ minutes from pouring, after which time it slowly increases to a maximum of 20° C at $7\frac{1}{2}$ minutes. As might be expected, during solidification the temperature differences are smaller towards the centre of the block than near the wall.



Casting TG2E (Fig.75).

The curves of temperature difference obtained during the solidification of casting TG2E are similar in most respects to the corresponding set obtained from casting TG2A.

Although the beginning of curve 3 was not recorded, the curves shows that after $2\frac{1}{4}$ minutes the temperature difference between position c and d (near the wall of the block) is 17°C , after which it slowly rises to 20°C at $3\frac{3}{4}$ minutes, and then decreases after about 5 to 6 minutes. Finally there is a gradual decrease, which is common to all three sets of curves and continues until room temperature is reached.

In conjunction with the temperature-difference measurements made during the solidification of castings TG2A and TG2E, direct temperature measurements were made at the central positions. It was again found, as in casting TGA, that the time taken to reach the peritectic temperature coincided with the maximum temperature differences recorded after $7\frac{1}{2}$ minutes and $5\frac{3}{4}$ minutes respectively.

Curves showing the actual temperature distribution existing during the solidification of casting TG2A and TG2E were prepared by subtracting (or adding) from the direct temperature curve (obtained on a potentiometer) the temperature difference values recorded by the differential couples. (These curves are not reproduced). It was found that the errors were cumulative,

being reproduced in successive calculations; in some cases, liquidus temperatures were found to be as much as 15-20°C below the true values. Similar curves representing the thermal distribution along the axis of the block were prepared, but they also were clearly unreliable and are likewise not included.

Although the numerical accuracy of the temperature difference curves is suspect, the general form of the curves must be basically correct. They are therefore of some value and are discussed on page 160 in relation to the results of direct measurement.

(2) Direct Method.

In order to follow the progress of solidification of the different types of castings, it was found necessary to interpret the thermal data in terms of a "corrected" iron-carbon equilibrium diagram. The first step consisted in "translating" the measurements recorded by the apparatus, in terms of direct units of temperature and time.

A method similar to that adopted for the calibration of the difference curves, was employed for the calibration of the direct temperature measurements. By passing various standard e.m.f.s. through the recorder, it was possible to calibrate directly the traces obtained in terms of e.m.f.s. These were then compared with calibration tables (for the type of thermocouple employed) to obtain the temperature in degrees Celsius. For this series of experiments it was, therefore,

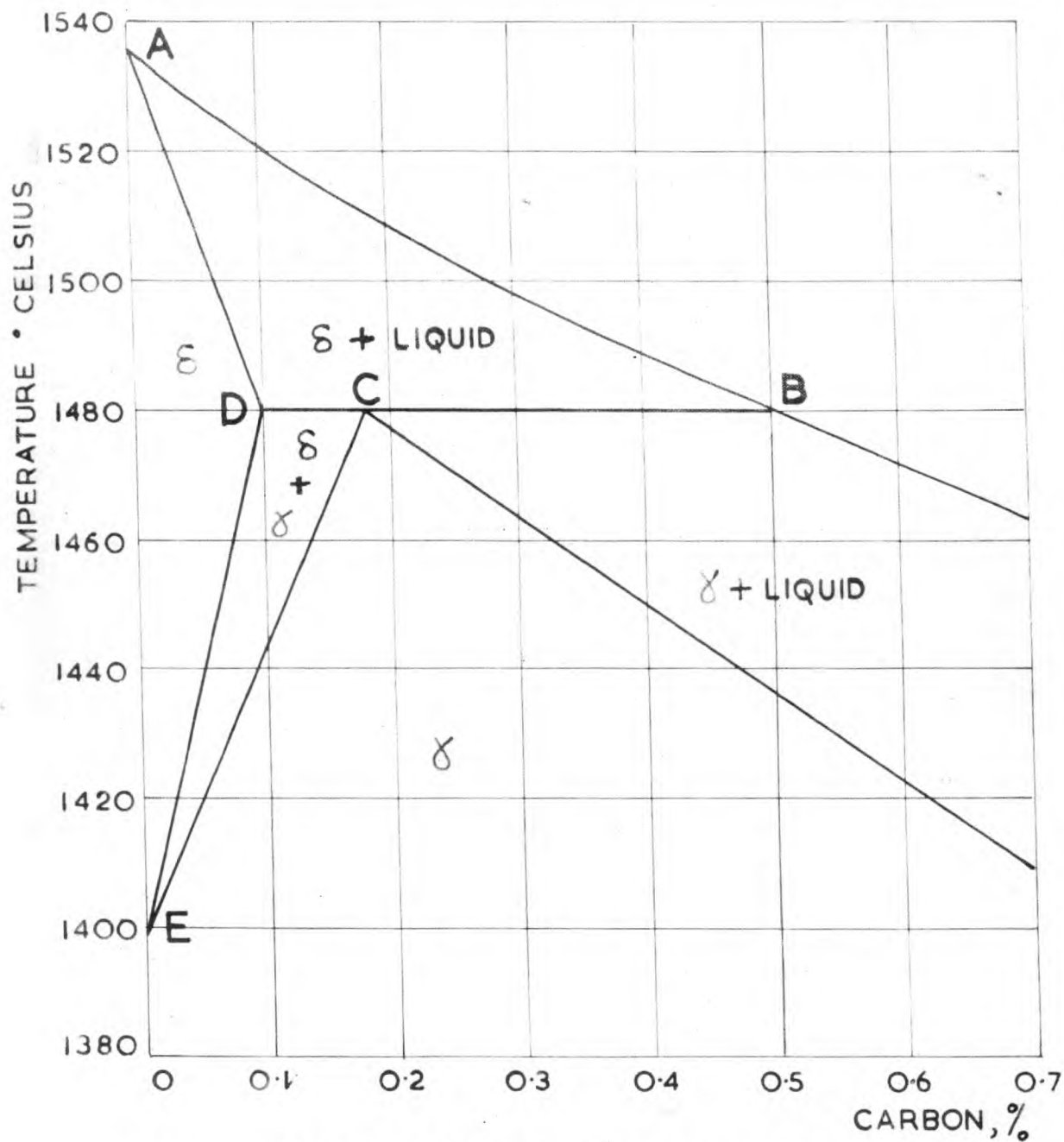


FIG.76 IRON-CARBON EQUILIBRIUM DIAGRAM

only necessary to construct scales representing deflection of the recorder needle in terms of e.m.f. in order to determine the temperature. The same scale was employed for calibrating the length of the trace in terms of time as that used for the temperature-difference method.

The iron-carbon equilibrium diagram shown in Fig. 76 has been "corrected" to allow for the manganese, silicon etc., present in the steel. Certain principal invariant points were taken from the diagram given in Metals Handbook⁽⁶⁵⁾; others are based on the results of West⁽¹⁶⁾ and the present investigation.

The liquidus boundary, ^BAE and AE produced, was constructed by means of the results of the present investigation and is in close agreement with the values reported by West. Although there was some variation in the manganese and silicon contents of the different castings (page 93), namely between 0.24% - 0.49% and 0.08% - 0.20% respectively, the relevant ternary equilibrium diagrams indicate that within these limits these elements have relatively little effect on the melting point. Hence, the liquidus in the "corrected" equilibrium diagram can be used with reasonable accuracy to interpret the various sets of cooling curves obtained during the investigation.

The solidus boundary, ^CDF, is probably the most doubtful curve in the diagram, because of experimental difficulty in determining the exact temperature at which

solidification is completed. Nevertheless, it is believed that the error is not likely to exceed a few degrees.

Let us consider the manner of solidification of a representative casting, containing 0.24% carbon, in the light of the equilibrium diagram (Fig.76).

At the liquidus temperature of approximately 1504°C , solidification would begin by crystallization of δ -iron. This would continue till the peritectic temperature (approximately 1480°C) was reached, where the δ -iron would be converted into γ -iron, some liquid remaining after the completion of the reaction. This liquid would solidify as γ -iron while the temperature fell, until the solidus was crossed at about 1472°C .

The cooling curves obtained during the solidification of the different castings can be readily interpreted in accordance with the above deduction from the equilibrium diagram. After filling the mould, there was an initial rapid fall in temperature (See for example Fig.77) to the liquidus temperature. While the temperature indicated by the couple placed nearest the wall (Curve 1) fell too rapidly past the liquidus to allow an arrest to be recorded, the temperature at the other positions remained constant for a time and then began to fall slowly as solidification proceeded across the casting. An increase in the rate of cooling occurred after the completion of the peritectic reaction at about 1480°C ;

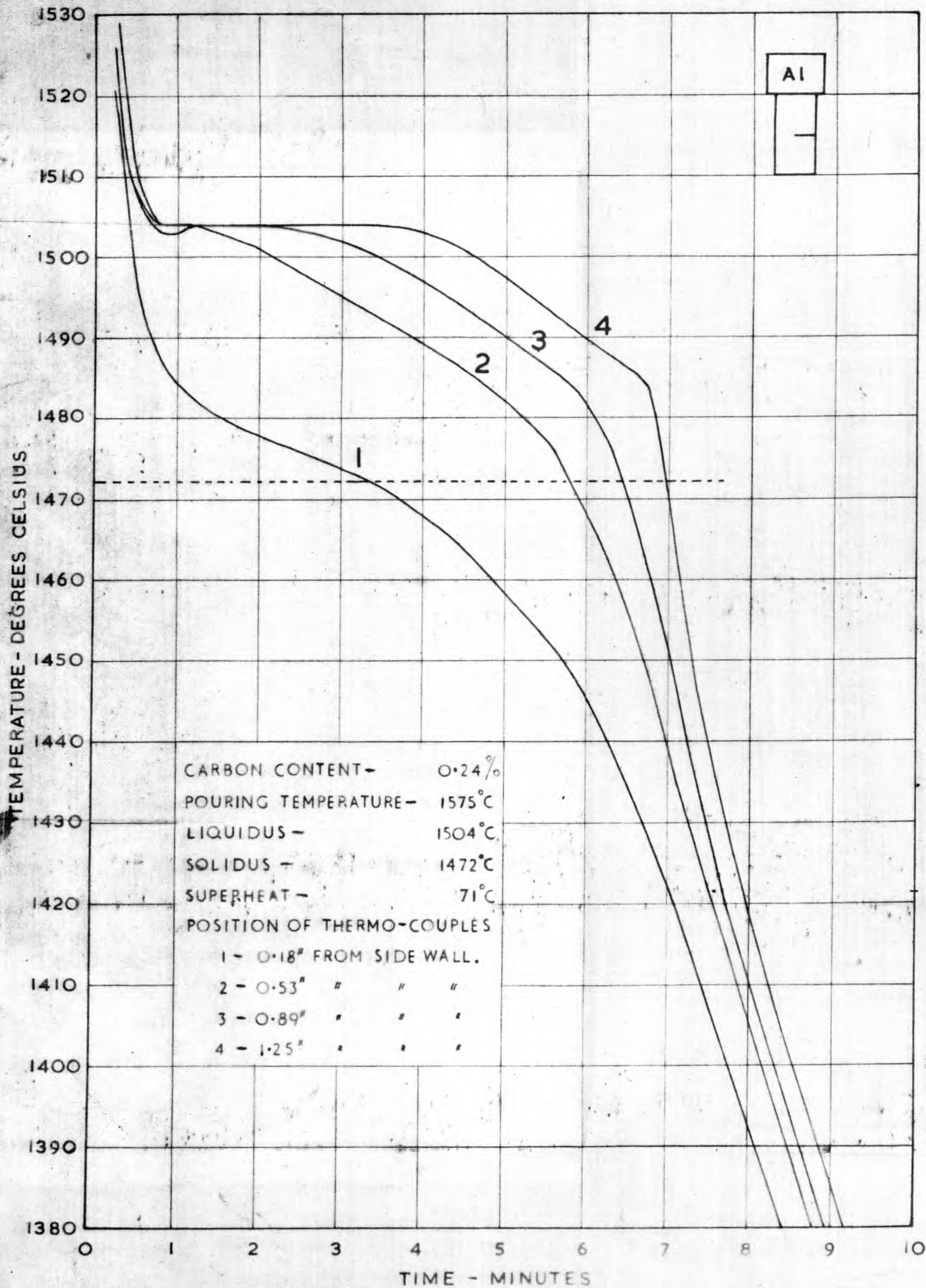


FIG. 77 COOLING CURVES OBTAINED AT POSITIONS ON TRANSVERSE SECTION OF TOP FED VERTICAL BLOCK - APPROX. 2.2" FROM HEAD.

this temperature was most clearly indicated by the couple at the centre of the casting. The temperature of the solidus cannot be obtained from the curves, so an "arbitrary solidus" temperature has to be taken from the equilibrium diagram. This value can be used to estimate the relative times of complete solidification at different positions in the block, and to compare one cast with another.

Casting Al. (Fig.77).

Cooling curves were obtained from thermocouples situated at four positions across the top-fed vertical block, approximately midway between the head and the bottom. The couples were spaced at intervals of approximately 0.36", one being 0.18" from the wall and another in the central axis of the block. It may be seen that within 45 seconds after pouring, the temperature of the metal across the block was at the liquidus value of 1504°C , except for the position nearest the wall. At this position the metal had already solidified without any measureable arrest. The couple situated on the axis of the block recorded an undercooling effect of approximately $1^{\circ} - 2^{\circ}\text{C}$.

The progress of solidification can be followed by noting the times when the temperature at the various positions began to fall below the horizontal arrest. Thermocouple 2, (0.53" from the wall) indicated that solidification started there after 1 minute 12 seconds. At position 3 (0.88" from the

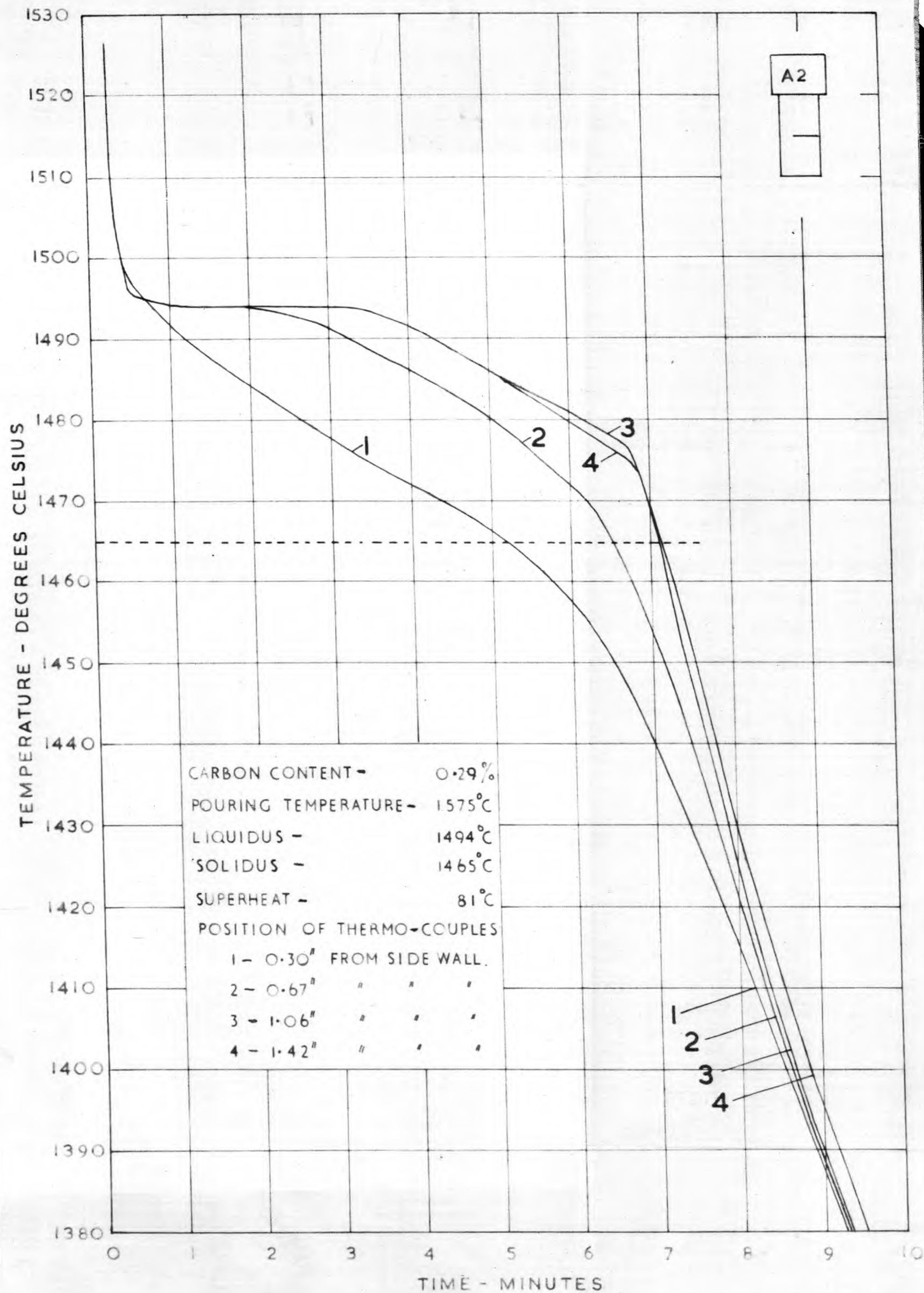


FIG.78 COOLING CURVES OBTAINED AT POSITIONS ON TRANSVERSE SECTION OF TOP FED VERTICAL BLOCK - APPROX. 2.3' FROM HEAD.

wall) solidification began after 2 minutes 15 seconds, and at position 4 (1.25" from the wall) after 3 minutes 30 seconds. The difference in the shape of the different cooling curves as they leave the liquidus arrest is, perhaps, related to the mode of crystallization. Curves 3 and 4 show a more gradual fall in temperature than curve 2 when leaving the arrest.

The "arbitrary solidus" for this casting (0.24% carbon) is given by Fig.76 (p. 118) as 1472°C , from which the following times of complete solidification across the block may be derived:-

0.18"	from wall	3 minutes 24 seconds
0.53"	" "	5 " 51 "
0.89"	" "	6 " 27 "
1.25"	" " (at centre)	6 " 57 "

This experiment was repeated three times and the results agreed very satisfactorily.

Casting A2. (Fig.78).

A similar set of cooling curves was obtained for a further four positions across the block, again approximately midway between the head and the bottom. The thermocouples were placed at intermediate position to those in casting A1, one being 0.3" from the side wall and the middle two equidistant from the central axis, 1.06" from either wall.

As would be expected, the cooling curves had the same form as those obtained from casting A1. There were minor

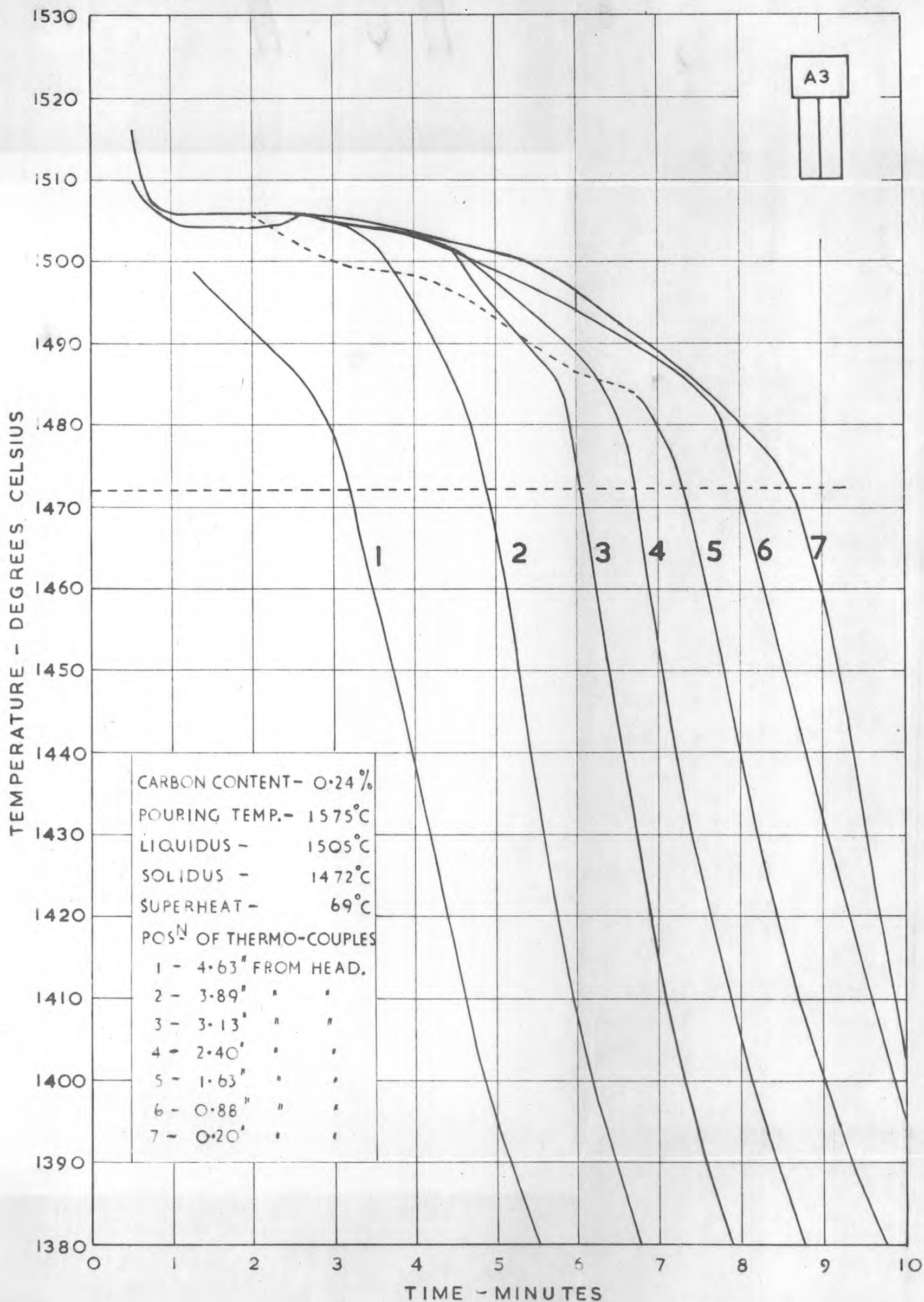


FIG.79 COOLING CURVES OBTAINED AT POSITIONS ON CENTRAL AXIS
OF TOP FED VERTICAL BLOCK.

differences between the two sets of curves, probably due to the difference in composition and superheat, so that an exact comparison of the actual times and temperatures cannot be made.

It may be seen that although the couple nearest the wall (Curve 1, Fig.78) does not show a horizontal arrest, the rate of cooling is considerably reduced at the liquidus temperature. The shape of curve 2 (0.67" from the wall) after the liquidus arrest is similar to that of curves 3 and 4 in casting A1. The curves for the positions equidistant from the central axis are very similar, the readings from both couples indicating that the liquidus arrest was completed at the same time; the rates of cooling are the same, and are similar to curve 2.

Taking an "arbitrary" solidus value of 1465°C , the times for complete solidification were:-

0.3" from wall	5 minutes 12 seconds
0.67" " "	6 " 30 "
1.06" " "	7 " 6 "
1.06" " other wall	7 " 6 "

Casting A3. (Fig.79).

A set of cooling curves was obtained for positions along the central axis, during the solidification of a vertical top-fed block. The casting conditions were similar to those for casting A1 and A2. One of the thermocouples was situated 0.37" from the bottom and another 0.20" from the base of the head; the remaining couples (five) were

placed at intervals of 0.75" along the axis.

The progress of solidification along the block can again be deduced from the form of the cooling curves, although there are some irregularities in this series. Within about 1 minute after pouring the metal had cooled rapidly to a liquidus temperature of 1506°C . Although the couple nearest the bottom did not record any temperature reading until 1 minute 15 seconds after pouring, it is probable that even if it had, no definite arrest would have been recorded, owing to the rapidity of solidification close to the wall. After $2\frac{1}{2}$ minutes the temperature of couple 2, 1.11" from the bottom, fell below the liquidus arrest temperature. The other curves, however, did not depart at more or less regular time intervals from the horizontal arrest temperature as in the series described above. Instead, after about 3 minutes, the temperature at all these positions fell slowly, at approximately the same rate, until about $4\frac{1}{2}$ minutes after pouring when the curves branched apart. This indicates that the rate of cooling was different at different positions, until the peritectic temperature of about 1480°C was reached.

Taking the "arbitrary solidus" as 1472°C , it may be seen that final solidification took place progressively from the bottom towards the top, the times being as follows:

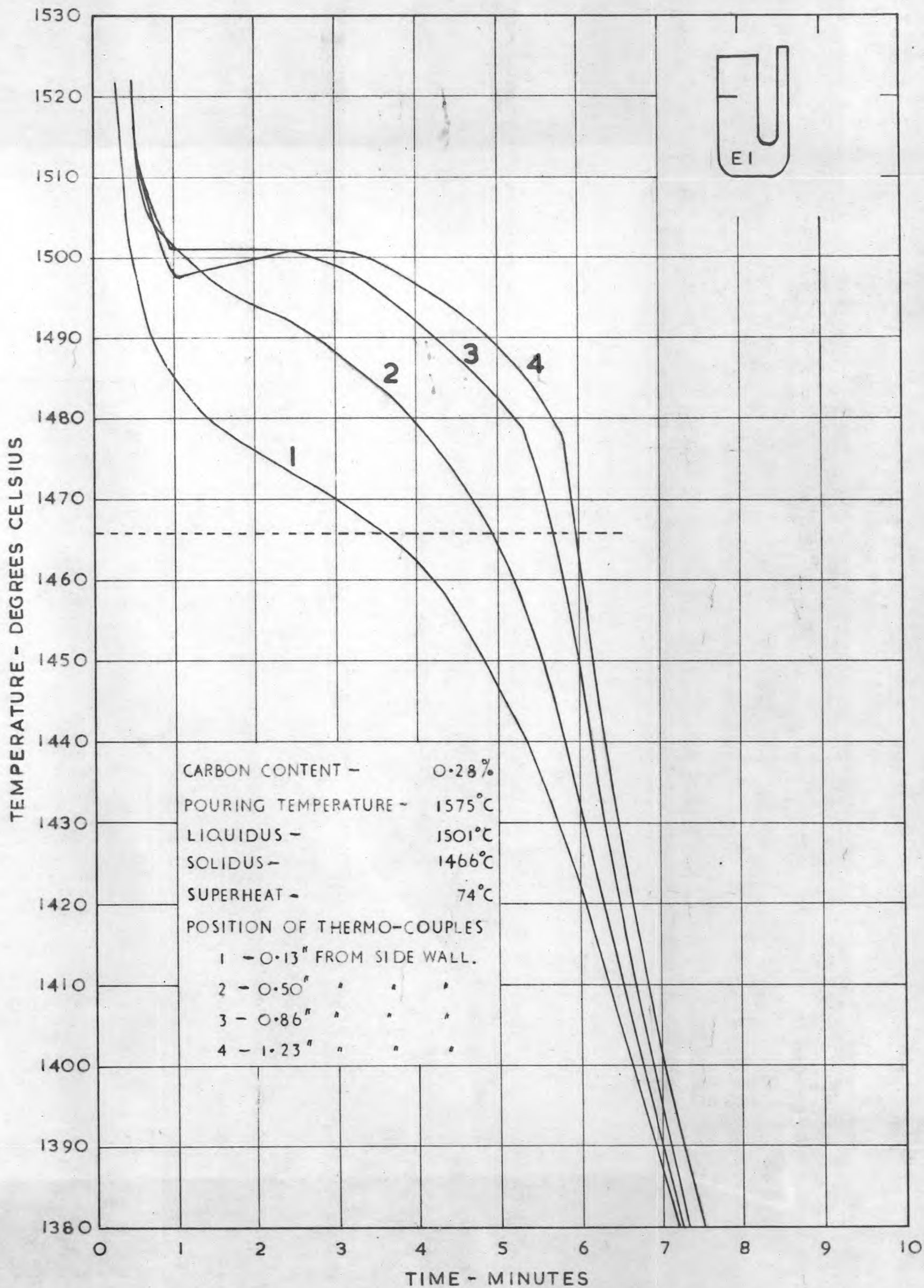


FIG.80 COOLING CURVES OBTAINED AT POSITIONS ON TRANSVERSE SECTION OF BOTTOM FED VERTICAL BLOCK - APPROX. 2.3" FROM HEAD.

<u>Distance from Head</u>	<u>Distance from Bottom Wall</u>	<u>Time.</u>
4.63"	0.37"	3 mins. 12 secs.
3.89"	1.11"	4 " 54 "
3.13"	1.87"	6 " 3 "
2.40"	2.60"	6 " 42 "
1.63"	3.37"	7 " 18 "
0.88"	4.12"	8 " 0 "
0.20"	4.80"	8 " 48 "

The anomalous fall in temperature in the early stages of curve 5 is probably due to a slag inclusion attaching itself to the silica sheath of the thermocouple, or to the formation of a cavity.

Casting El. (Fig.80).

Cooling curves were obtained from a series of thermocouples situated across a bottom-fed vertical block, approximately halfway up the casting. Four couples were employed, spaced at intervals of 0.36". One was 0.13" from the wall and another was on the vertical axis of the block. Although the thermocouples were placed in similar positions to those in casting Al, the cooling curves obtained indicate certain important differences. The thermocouples situated at 0.13" and 0.50" from the side wall did not record a horizontal arrest temperature, whereas in casting Al, only the couple placed at 0.18" from the edge failed to show an arrest. Furthermore, in the bottom-fed castings the rate of cooling

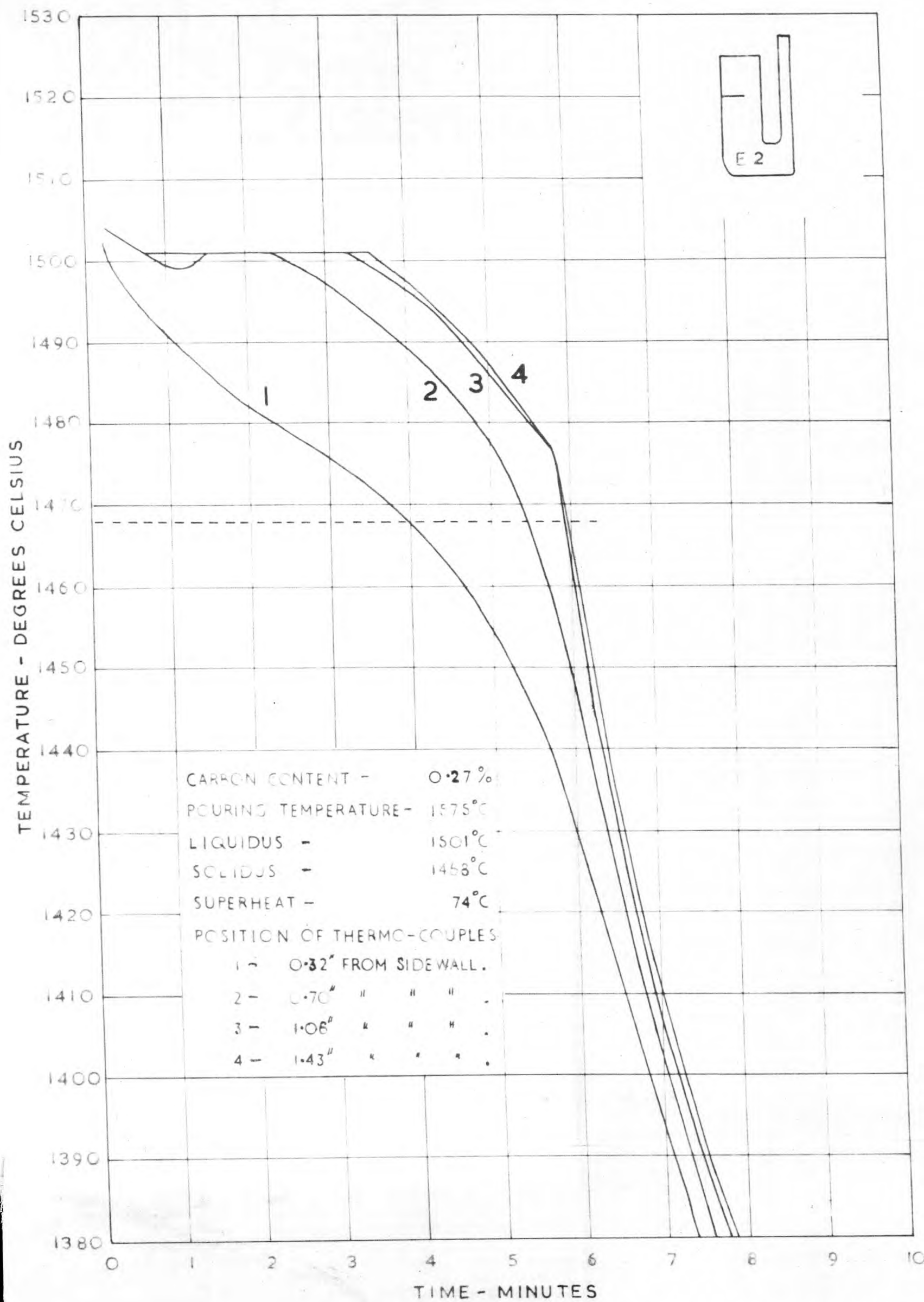


FIG.81 COOLING CURVES OBTAINED AT POSITIONS ON TRANSVERSE SECTION OF BOTTOM FED VERTICAL BLOCK - APPROX. 2.4' FROM HEAD.

for the 0.50" position was faster than for the corresponding position in the top fed block.

Couple 3 (0.86" from the wall) showed an arrest lasting $2\frac{1}{2}$ minutes, and couple 4 at the centre of the block, after undercooling by $2^{\circ} - 3^{\circ}\text{C}$, showed an arrest lasting 3 minutes. These times are both shorter than for the corresponding positions in casting A1.

Assuming an "arbitrary solidus" of 1466°C , solidification was completed at the following times:-

0.13"	from the wall	3 minutes 36 seconds
0.50"	" " "	4 " 54 "
0.86"	" " "	5 " 42 "
1.23"	" " "	5 " 57 "

Compared with the corresponding times of complete solidification for casting A1, the above evidence indicates a faster rate of solidification. For example, solidification at the centre was completed 1 minute earlier in the bottom-fed type of block than in the top-fed type.

Casting E2. (Fig.81).

A further series of cooling curves was obtained from couples across the block approximately halfway between the head and the top face. They were situated at intermediate positions to those in casting E1, one being 0.32" from the side wall and the middle two being equidistant from the central axis. This arrangement was similar to that employed for casting A2.

The cooling curves obtained were basically similar

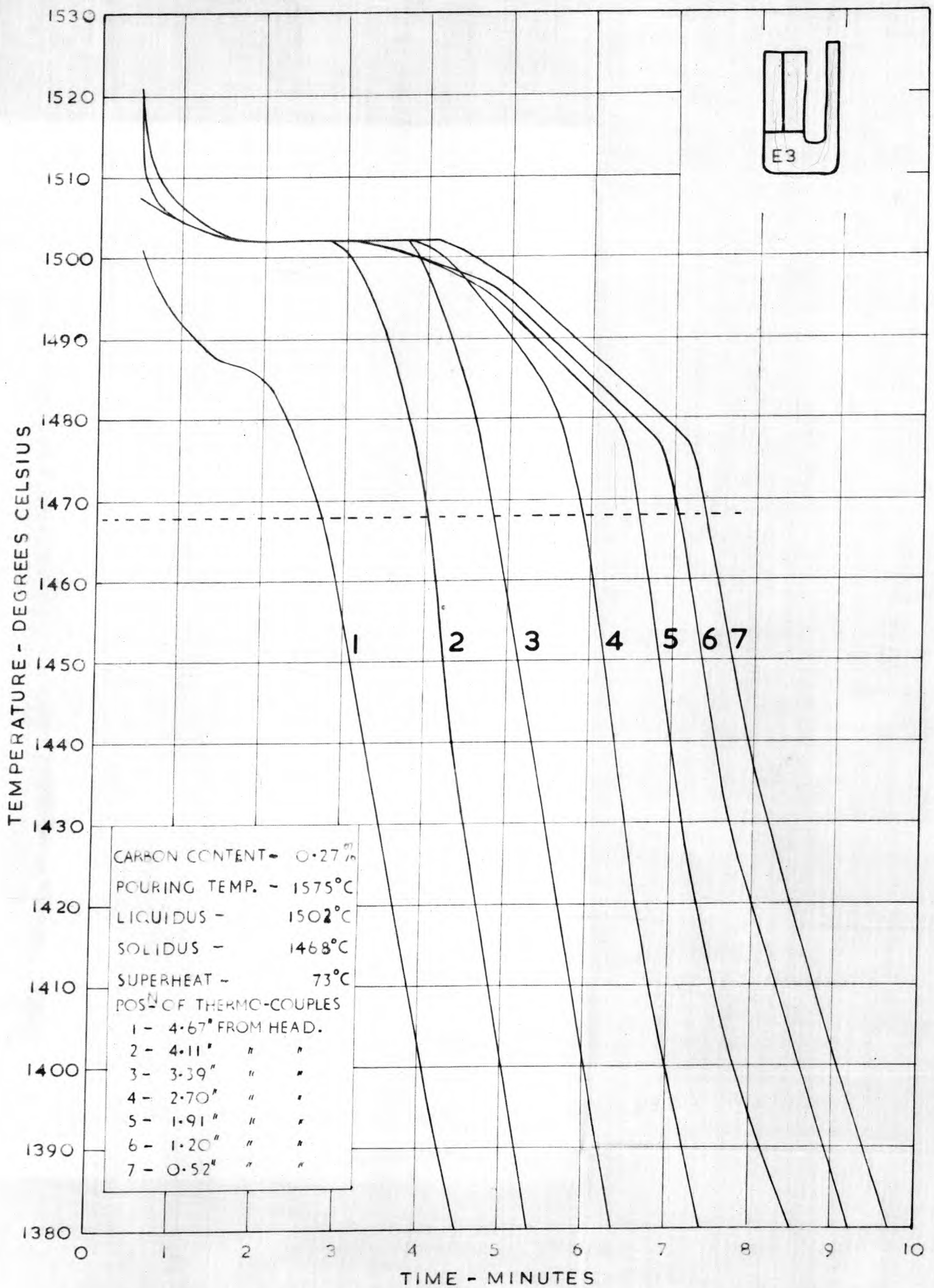


FIG.82 COOLING CURVES OBTAINED AT POSITIONS ON CENTRAL AXIS
OF BOTTOM FED VERTICAL BLOCK.

to those of casting El. Owing to rapid solidification from the walls, no definite arrest was recorded by the couple placed 0.32" from the wall, and as might be expected from the position of the couple, the form of the cooling curve was approximately intermediate between those of curves 1 and 2 in casting El. After undercooling about 2°C , the temperature of the couple situated 0.70" from the wall fell below the liquidus arrest $2\frac{1}{2}$ minutes after pouring. The temperature of the couples situated equidistant from the central axis left the liquidus arrest almost simultaneously, i.e., after $3\frac{1}{2}$ and $3\frac{1}{2}$ minutes respectively. In all cases, the rates of cooling were similar to those found in casting El.

Assuming an "arbitrary solidus" value of 1468°C , the times of complete solidification were:-

0.32" from wall	3 minutes 57 seconds
0.70" " "	5 " 27 "
1.06" " "	5 " 54 "
1.06" " other wall	5 " 57 "

These times indicate a faster rate of solidification than for the corresponding positions in the top-fed casting. This is in agreement with the comparison made between the solidification times of castings Al and El.

Casting E3. (Fig. 82).

A similar set of cooling curves was obtained during the solidification period, along the central axis of a bottom-fed vertically-cast block. One couple was placed 0.33" from the top wall (4.67" from the head) and another

0.52" from the feeder head; the remaining (five) couples were placed at 0.75" intervals along the axis.

All the couples with the exception of the one nearest the top wall, recorded a liquidus arrest within $1\frac{1}{2}$ minutes after pouring. It may be seen that close to the top wall, solidification was so rapid that the arrest was not recorded.

After $2\frac{3}{4}$ minutes the temperature of the couple placed 4.11" from the feeder head (0.89" from the top wall) fell below the liquidus arrest, followed later by the other couples along the axis. Solidification began in some of the lower positions before those higher up in the block, and for this reason the liquidus arrest at the different positions was not completed in regular succession along the block towards the feeder head. However, if the "arbitrary solidus" value is taken as 1468°C , it appears that the completion of solidification progressed continuously towards the feeder head from the end wall. The times of complete solidification were:-

<u>Distance from Head.</u>	<u>Distance from End Wall</u>	<u>Time.</u>
4.67"	0.53"	2 min. 39 secs.
4.11"	0.89"	4 " 0 "
3.39"	1.61"	4 " 48 "
2.70"	2.30"	5 " 54 "
1.91"	3.09"	6 " 33 "
1.20"	3.80"	7 " 3 "
0.52"	4.48"	7 " 24 "

The results confirm those of casting E1 in showing that the total solidification time of the bottom-fed casting was 1 minute less than that for the casting fed at the top

(casting A3) (page 124).

(d) "Pour-out" Tests.

As was stated earlier, top-fed vertical castings were used for the "pour-out" experiments. They were bled at intervals of one, two, three and four minutes after pouring, and show the approximate extent of solidification at these times.

The casting bled after one minute (Fig.83) consists of a shell of metal approximately 0.40" thick, which had solidified around the mould wall in a columnar type of crystallization. The solidified shell had retained the contour of the mould wall completely, including that of the bottom face. The interior surface of the casting was relatively smooth.

The casting bled after two minutes (Fig.84) showed that solidification had progressed to a depth of approximately 0.6". Although the structure was mainly columnar there were small areas of nuclear or fine equiaxial crystals adhering to the inner extremity of the columnar crystals. This gave the interior of the casting a roughened surface. At the bottom of the block there was a layer of equiaxial-nuclear crystals, superimposed on a thin layer of columnar crystals, to a total depth of about 0.85".

The casting bled after three minutes (Fig 85), showed that columnar crystallization had ceased after two minutes, changing to a nuclear or fine equiaxial crystallization

The columnar and fine equiaxial/nuclear crystals had a total thickness of about 0.75". One inch of metal had solidified at the bottom of the block.

The structure of the casting bled after four minutes (Fig. 86) indicates that a mass of partly solidified metal had fallen towards the head, leaving several cavities in the central portion of the block. There was no solidification front such as was found in the previous castings. From this structure it would appear that the mixed equiaxial/nuclear crystallization had progressed, though by no means to completion, throughout the whole central portion from numerous independent nuclei. If the block had been left to solidify under normal conditions, the crystallites would have multiplied and developed until the whole mass was solid, resulting in the structure shown in casting WOX6A. (Figs. 52, 57).

The "pour-out" results, however, do not give a true picture of the progress of solidification because of the presence within the crystallites of entrapped liquid which solidified after the casting had been bled.

The results are considered in relation to those of the corresponding temperature determinations in the next section of the thesis.

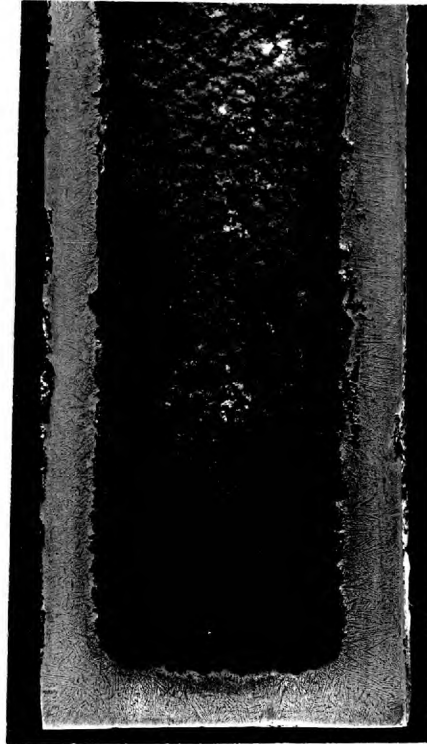


Fig. 83. Bled after 1 Min.



Fig. 84. Bled after 2 Mins.

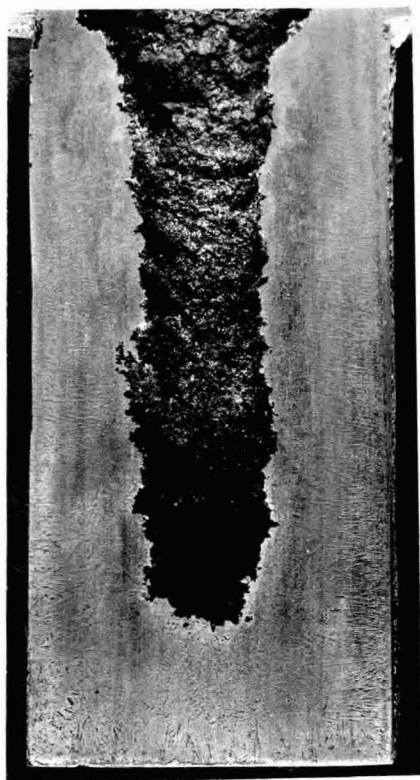


Fig. 85. Bled after 3 Mins.



Fig. 86. Bled after 4 Mins.

SECTION FOUR.

DISCUSSION.

1. Introduction.

It was shown in section one of the thesis that the mechanism of solidification of metals and alloys has been the subject of many investigations, and although certain basic principles have been firmly established, and the main controlling factors determined, there remain many aspects of the subject requiring more detailed study.

In outline, the solidification of any metal or alloy can be regarded as a two-stage process in which there is an initial nucleation of a solid phase from a liquid, followed by a period of growth and development, leading to the elimination of the liquid phase.

The factors which control the mode of solidification may be broadly divided into two main groups:- (a) Certain inherent and invariant properties of the metal or alloy, e.g., crystallization habit, freezing point or freezing range (in the case of alloys the phase relationship between the liquid and solid as defined by the relevant phase diagram), specific heat, thermal conductivity, latent heat of freezing, and shrinkage. (b) Certain extraneous factors such as variations in the thermal conditions, ~~determined by~~ mould materials, mould design, casting temperature and to a certain extent by the influence of nucleating agents and impurities.

The results of previous investigations of the

solidification of steel castings have clearly shown that the thermal conditions have important effects on the structure and the rate of solidification, but the particular effects of the individual thermal factors has not always been fully described and distinguished. In some cases the factors influencing the thermal conditions and hence the solidification process may act in opposition. For example, as a general rule an increase in the casting temperature (degree of superheating) favours increased columnar development and an increase in the time of solidification. However, other factors, which themselves influence the thermal conditions, may act in opposition to the normal effect of casting temperature. Thus, a sand casting with a large feeder head, poured at a relatively low casting temperature (cf. casting Y1 Fig. 91), between pages 139-140 may show the same type of structure as one with a smaller head but poured at a higher casting temperature. (cf. casting SHA2, Fig. 63, between page 108-109. The method of feeding is another factor which affects the mode of solidification. A bottom-fed casting, for example, has a different crystal structure from a top-fed casting poured with the same degree of superheating.

Correct differentiation between the various factors is therefore essential in studying the mechanism of solidification. The significance and importance of these factors is considered in detail below in the light of the results of the present investigation.

2. Superheat and Casting Temperature.

The results of previous investigations^(42,16,64) have shown that although the superheat of molten metal is rapidly lost when the metal is poured into a mould, it nevertheless has a profound effect on the nucleation and rate of solidification of the metal, and hence on the final crystal structure. In the present investigation it has been shown that by systematically increasing the casting temperature a series of structures of increasing coarseness is obtained (Figs. 62 - 71, facing p.109), ranging from a fine nuclear structure with zero superheat, to equiaxial dendritic crystals and pronounced columnar wall growth for higher casting temperatures.

A consideration of Tammann's original laws of nucleation would suggest that the fine nuclear structure, produced by a low casting temperature, was the result of marked undercooling of the liquid metal prior to nucleation. Subsequent work, however, has shown that relatively large masses of metal do not undercool to any marked extent. The greatest degree of undercooling reported in the literature for a mass of steel of about the size of the present castings, is 9°C ; for smaller experimental melts, undercooling of the order of $5^{\circ} - 13^{\circ}\text{C}$ has been reported for steel, depending on the degree of superheat. As a general rule, the greater the superheat the greater the degree of undercooling. In the

present investigation the maximum undercooling observed was of the order of 3°C for a superheat of 75°C (Fig.80). It appears, therefore, that the crystal structures of castings cannot be adequately explained in terms of undercooling phenomena only; in fact the experimental evidence suggests that if undercooling is an important factor, its effect is the opposite of what one would deduce from Tammann's law, since the greater the degree of undercooling the coarser is the structure obtained (i.e., the smaller the number of nuclei formed).

One suggested explanation of the way in which solidification is started is that nucleation is initiated by "pre-existing nuclei", such as sand and oxide particles suspended in the liquid. Thermodynamic considerations have also indicated the possible existence of certain areas of "long range atomic order", which are retained in the liquid state after melting and may initiate nucleation on cooling (see page 25). In terms of this theory, it is possible to account for some features of the structures of castings. For example, the lower the superheat the more numerous are the "pre-existing" nuclei. Thus, should a casting be made from metal which has been melted and poured without being superheated, a large number of nuclei will be retained and consequently a very fine type of nuclear structure will result (cf. Casting SHAL Fig.62, facing page 109). If, however, the melt is superheated, the areas of "long range

atomic order" will be gradually eliminated and the sand or oxide particles dispersed or taken into solution. During subsequent cooling and solidification in the mould the relatively small number of "pre-existing nuclei" will result in a relatively coarse structure. The results of the present investigation (Figs. 62-66) support this theory. Thus, as the superheat is raised, and the number of nuclei diminished, there is a transition from a fine undeveloped nuclear structure to a well-developed dendritic form, typified by casting SHA6, Fig. 66. Possibly, however, if the metal were superheated to a sufficiently high temperature so that all the "pre-existing" nuclei were destroyed, and more pronounced undercooling might be caused, and a fine structure might result.

The formation of columnar crystals and increase in their length, as a result of an increase in superheat, is not considered to be due to the removal of nuclei etc., but to a change in the nature of the thermal conditions existing between the solidifying metal and the mould. This is discussed later on page 175.

3. Convection currents and Mould Arrangement.

Interest in the possible influence of convection currents was first aroused by Gray⁽⁵⁴⁾, who found that widely different crystal structures were obtained upon varying the position of the feeder head and the arrangement of the block.

This, he considered to be due to the effect of, or absence of, convection currents in the liquid metal. For example, in a horizontal block where feeding was from one end, convection currents were considered to be almost absent, owing to the small vertical height. In a vertical block which was top fed and bottom run the currents would be more pronounced, since hot steel which was in the bottom of the mould when casting was completed, would rise to the top when the turbulence had ceased. Convection currents were considered to be strongest in a casting which was bottom poured and bottom fed; hot metal from the bottom of the casting and the head would continue to rise to the top until freezing was almost completed. It was suggested that this action would delay solidification and result in a coarse confused dendritic crystal structure.

As was stated on page 33, Schaumn⁽⁶¹⁾ found, by direct temperature measurements, that if grey cast iron was sufficiently superheated, a vertical casting could be inverted three times after the metal had been poured, each inversion being followed by a movement of hot metal from the bottom of the casting towards the top. The movement of the hot metal ceased when the liquidus temperature was reached. Schaumn considered that the temperature changes following the inversion of the casting were due to the action of convection currents. Although it has been suggested that the temperature reversals might be the result of turbulence, the similarity between the

form of the temperature curves following the inversion of the casting and that obtained immediately after pouring, seem to indicate that the temperature changes are due to the same effect, i.e. convection currents.

Although Schaumann's work indicates that convection currents do arise in vertical castings, there is doubt as to their actual influence on the process of solidification. However, an examination of the structures of the series of castings described in Section Two Figs. 2,5,8,11,15, suggests that the promotion of convection currents by different mould arrangements resulted in a marked change of crystallization. In casting A1, Fig.2, which was designed to reduce convection currents to a minimum by placing the feeder head vertically over the block, a fine nuclear structure was obtained, whereas, in casting E1, Fig.15, in which convection currents were increased to a maximum by placing the hot source of metal vertically below the block, a coarse dendritic structure resulted. Casting E1, C1 and D1, Figs.5,8 and 11, which were arranged at intermediate "positions", and therefore subjected to increasing convective effects, had transitional structures between the nuclear form of A1 and the dendritic form of E1. As all other conditions, such as casting temperature and composition, were the same, it might seem reasonable to conclude therefore that the effect of convection currents was to reduce the nucleation rate, thus providing time for the crystallites to develop dendritically before being restrained

by the growth of neighbouring crystals.

It must be borne in mind, however, that the position of the feeder head will have an effect on the number of "free crystallites" settling through the block towards the bottom. If the head is at the top and is relatively small, so that it begins to freeze over before solidification is far advanced in the block, free crystallites may be expected to sink from the head into the block, adding to any that may develop in the block itself; new crystallites will tend to interfere with dendritic growth, columnar or equiaxial, in the block, and therefore favour the formation of a nuclear structure. On the other hand, where the head is at the bottom, the only free crystallites available to affect the structure will be those formed in the block, and many of them will be able to settle out of the block into the feeder at the bottom.

The intermediate castings (B1, C1 and D1) showed a progressive transition from the nuclear structure of A1 to the dendritic structure of E1. This might be ascribed to the effect of increasing convection currents, though it is difficult to see how such currents could produce these effects. It seems more reasonable to suppose that the progressive lowering of the feeder head, and changing inclination of the block, resulted in a progressive transition from a condition favouring the feeding of crystallites from the head into the block to one favouring the escape of crystallites from the

block into the head.

Before passing on to the consideration of the next series of castings, it may be pointed out that while no clear evidence has been obtained in this investigation that convection currents have any significant effect on the solidification process or the resulting microstructure of the castings, the results do not prove conclusively that convection currents are non-existent or have no effect.

In the second series of blocks (Figs.19,22,28 and 31), which were cast in order to avoid the side-wall effects, the structures obtained were not, in all cases, similar to the corresponding castings of the first series. In fact, the structure of the top-fed casting A2 (Fig.19) was dendritic whereas that of A1 was nuclear, and the structure of E2 (Fig.31) was mainly nuclear whereas that of E1 was dendritic. The structure of the intermediate castings B2,C2 and D2, were similar in most respects to those of the corresponding members of the previous series.

It was intended that the two series of blocks should be cast under exactly the same conditions, but unfortunately, the feeder heads of three castings in the second series were increased in size in order to prevent the formation of pipes, which had been present to some extent in the first series. Moreover, a quantity of exothermic compound was added to the heads. Both these modifications would increase the thermal capacity of the heads and might therefore

be expected to have an effect similar to that of a higher casting temperature, partly because a greater store of heat would be available for conduction into the block, and partly because as the block cooled and contracted, hot metal would be drawn by it from the large reservoir of molten metal in the head. Moreover, as the head would remain molten till a late stage in the solidification of the block, it would not feed solid crystallites into the block, even in the case of the top-fed vertical casting. For these reasons one would expect the second series of castings to have coarser structures than the first series, with longer columnar crystals growing from the walls, and equiaxial rather than nuclear crystals in the interior. Comparison of the two series of castings shows that the variations in their structures are in general in accordance with the above reasoning. Casting E2 is the outstanding exception, but as was stated earlier its structure is so anomalous that it seems necessary to reject it.

4. Size Effect.

The structure of the series of small blocks cast in the laboratory, were basically in agreement with those of the two larger commercial series. (These structures are described in detail on Pages 97-104, Figs.52-61). The small vertical top-fed block A (Fig.52), consisted of a columnar shell surrounding a central area of nuclear and equiaxial crystals - a structure that is intermediate between those of

castings A1 (Fig.2) and A2 (Fig.19). It appears, therefore, that in the small casting A, the heat content of the head relative to that of the block was greater than in the large casting A1 but less than in casting A2.

The vertical bottom fed block (Fig.56) consisted completely of columnar dendrites which have developed at right angles to the mould wall. This type of structure, with the absence of equiaxial crystals, has the appearance of a typical chill cast structure, and indeed temperature measurements (see tabulated data on pages 124 and 127) indicate that bottom-fed castings solidify more rapidly than similar top-fed castings. However, the difference between the small bottom-fed block (WX3, Fig.56) and the large block of similar type (E1, Fig.15) is not difficult to understand. If the width of casting WX3 was progressively increased from $2\frac{1}{2}" \times 3"$ towards that of the large casting ($4" \times 6"$), one would not expect the length of the columnar crystals to increase indefinitely so as always to meet at the centre line. When the block width exceeded a certain value, equiaxial crystals would begin to appear in the interior. [As the length of the columnar crystals in the large block (E1) is $1\frac{1}{4}" - 1\frac{1}{2}"$, and the casting temperature was about the same as for the small casting, it would seem that the "critical" maximum width for the production of a wholly columnar structure in this type of casting would be not much more than $2\frac{1}{2}"$]

An investigation of the effect of variation in the size and thermal content of the feeder head on the solidification of vertically cast top-fed blocks was made by Reynolds⁽⁶⁶⁾. Two series of castings were made. All the castings were poured under identical conditions, but the dimensions of the heads were varied both vertically and laterally from block to block. The dimensions of these two series of experimental blocks were the same as those of the castings in the present work, viz., 12" x 6" x 4" and 5" x 3" x 2½". The casting temperature was also the same, so a useful comparison of the structures may be made. The structures are shown in Figs. 87 and 88, together with particulars of the approximate weights of the heads and blocks.

Reynold's results show that although the casting temperature and position of the head remained unchanged, the size of the head had a marked effect on the crystallization, particularly in the smaller series.

An increase in the vertical height of the head caused a change of structure in the block from a mainly nuclear form, with a thin columnar shell (as shown by castings X1 and Y1, Fig. 87) to an equiaxial dendritic structure enclosed by a thick columnar shell (represented by castings X3 and Y3. This transition is similar to that caused by an increase in casting temperature, p.131.

The effect of an increase in the diameter of the head on the crystal structure however is less pronounced.

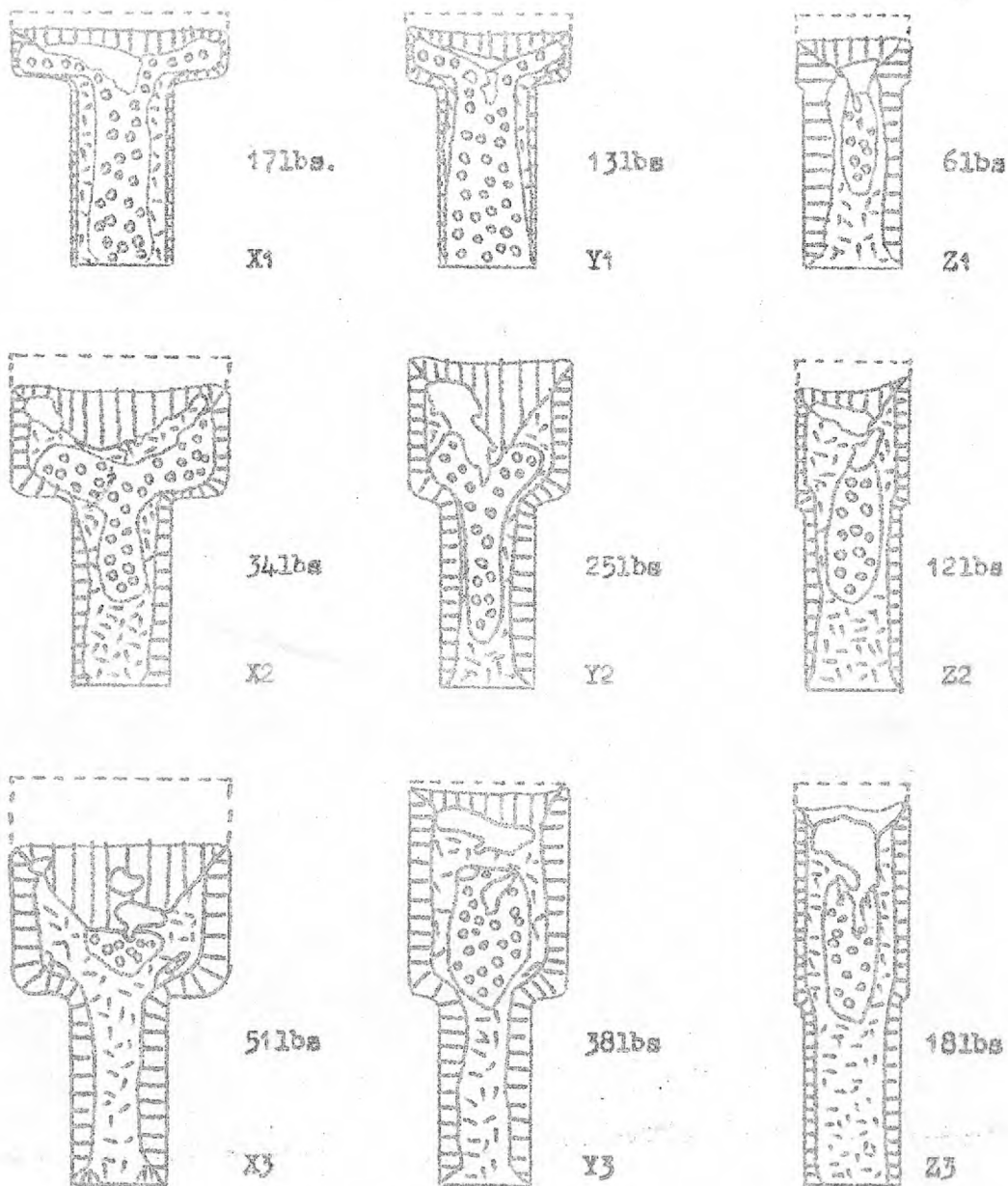


Fig. 87. Effect of Size of Feeder Head on Macrostructure of Top Fed Vertical Sand Castings
 (Size of Block 5" X 3" X 2½".
 Weight of Block alone 10lbs.
 Weight of Heads indicated.)

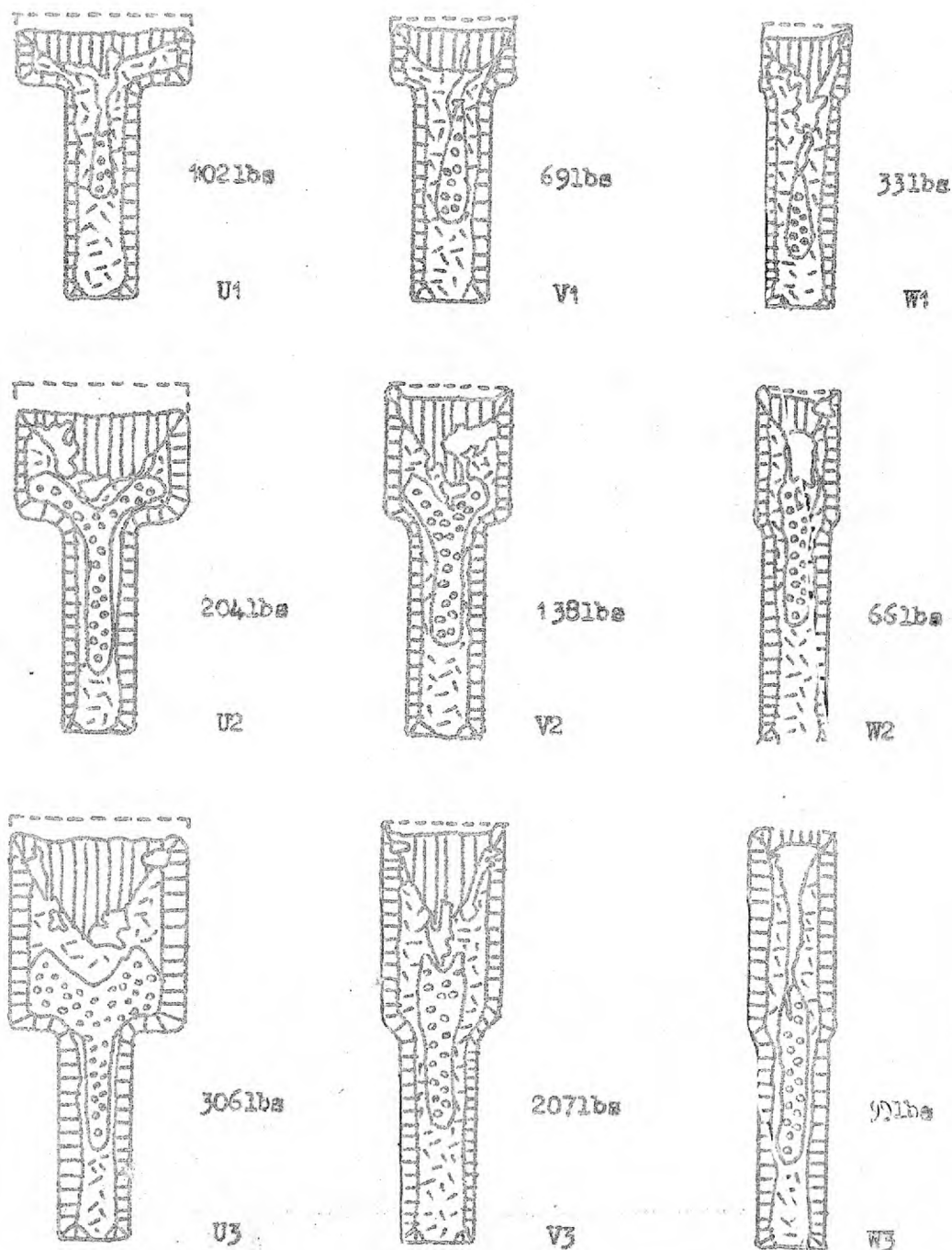


Fig. 88. Effect of Size of Feeder Head on Macrostructure of Top Fed Vertical Sand Castings.
 (Size of Block 12"x6"x14")
 Weight of Block alone 811bs.
 Weight of Head indicated.)

This, as explained below, is probably due to the greater cooling area (surface area) per unit volume obtained by increasing the diameter of the head (cf. castings X1 and Y1, and also X3, Y3 and Z3).

Examination of the larger series of castings (fig. 88) shows that the crystallization in the block was practically unaffected by an increase in the size of the head, either vertically or laterally. The structure consists, basically, of a columnar shell containing a central area of nuclear and equiaxial crystals, with the equiaxial crystals occupying the lower portion of the block. The extent of the upper nuclear zone varied slightly with the size of the head.

It is reported by Ruddle⁽⁹⁾ that Chvorinov⁽²⁰⁾, Janzo⁽⁶⁷⁾ and Caine⁽⁶⁸⁾ have considered the ratio of the surface area to the volume of a casting to be significant in assessing the solidification characteristics. The ratio is known as the "cooling factor". If the "cooling factor" is large (i.e. large cooling surface) then cooling is correspondingly rapid; whereas, if it is small cooling and solidification are slower. On this basis it is possible to deduce the probable influence of the size of the head on the crystal structures of Reynolds' castings.

Consider the castings to be divided into two parts: the feeder head and the block. If the "cooling factor" is determined for the two sizes of block, it is found that the small block corresponds to a value of 1.67, whereas, the

large block has a value of 0.92. Consequently the larger block, by virtue of its dimensions, should take longer to solidify than the smaller. This would be expected to result in coarser, dendritic crystallization. However, by fitting heads of different dimensions to these blocks, the solidification rates have been correspondingly altered. In the case of the smaller series of castings, the "cooling factor" varies from 1.86 to 1.04, and in the larger series from 0.96 to 0.54. These variations possibly reflect the differences in structure obtained throughout the two series of castings. For example, in the case of casting X1, with a "cooling factor" of 1.32, a finer structure is obtained than in casting X2, where the "cooling factor" is less (1.09). Similarly, the structure of casting X2 is finer than that of casting Y3, which has a "cooling factor" of 1.01. Since the "cooling factors" of the larger series of castings are less than for the smaller series, a coarser crystal structure is to be expected. This is probably correct, but without photographs of the macrostructures it is difficult to verify this observation. A comparison between the "cooling factors" of the large and small series also indicates that below a certain value, approximately 1, a decrease in the rate of cooling has only a limited effect in the resultant crystal structure.

[It is interesting to observe that if the "cooling factor" is determined for casting Z1, the value obtained

(i.e. 1.86) is larger than that for the block alone (i.e. 1.67). The effect of the head, therefore, has been to increase the rate of cooling of the casting, possibly accounting for the apparently anomalous crystal structure.)

It is also of interest to compare the structures of the castings made in the course of the present investigation with those made by Reynolds. Certain unexpected difficulties can be observed for example, the structures of castings A1 (Fig.2) and A2 (Fig.19) are different from those of Reynolds' corresponding block of similar dimensions (Fig.88, casting V1). Although the size of the head (60 lbs) of the nuclear casting A1 was similar to that of casting V1 (68 lbs) (Fig.88), the structure was very different. Similarly, although the size of the head (122 lbs) of the completely dendritic casting A2 approximates to that of casting V2 (138 lbs) (Fig.88), the structure of the latter is definitely finer and contains a relatively large region of undeveloped nuclear crystals. However, on the basis of apparently small experimental differences, it is possible to give satisfactory explanations of these anomalies. The heads of castings A2 and V2 are similar in size, but 1 lb. of exothermic compound was added to the head of casting A2, immediately after pouring. It is possible that the effect of the exothermic addition has been to eliminate, or greatly reduce, the feeding of "crystallites" from the head into the block, thus allowing the crystals in the block to develop dendritically. Although this addition

probably delayed the solidification of the head, it need not necessarily have delayed the solidification of the block appreciably.

In the case of castings Al (Fig.2) and Vl (Fig.88) although the heads were of similar size they were run in a different manner. In the top-run casting Al, the metal entering the mould was probably chilled to such an extent as to promote spontaneous nucleation. It is possible also that as the head cooled solid crystallites were fed from the head into the block. In the bottom fed casting however, such as Vl, the metal initially entering the mould was likewise chilled but this was displaced into the head by fresh hot metal entering the mould at a later stage. This effect has probably prevented spontaneous nucleation within the block during the early stages of solidification, thus permitting the crystals to develop dendritically. Later however, "crystallites" which have settled from the head have probably hindered the dendritic growth with the result that a mixed equiaxial/nuclear crystallization is obtained.

Gray⁽⁷⁶⁾ has found that the structure of a large top fed cylindrical casting (measuring approximately 42" x 6" diam.) was largely nuclear if the head was severely chilled by the insertion of heavy steel chills. It is considered that the mode of solidification was similar to that described for casting Al, that is to say, it involved the settling of solid crystallites from the head into the

The structure of a similar block cast from the same melt under exactly the same conditions of temperature but without chilling of the head, consisted of the normal mixed columnar, equiaxial and nuclear structure.

From the above discussion it may be concluded that although the size of a casting has an important influence on the rate of solidification, and hence the resultant form of crystallization, this is also largely affected by such aspects as the method of running, feeding and the ratio of the surface area to the volume of the casting.

5. THE MECHANISM OF SOLIDIFICATION.

(a) Theories of Solidification.

It is of interest to consider in detail the mechanism of solidification of sand castings and to attempt to develop a general theory in terms of which the effects of individual factors previously discussed may be explained. In the following discussion some of the main theories suggested by previous workers are considered in the light of the present results.

Heggie(32) has suggested that solidification occurs in the following manner:-

Immediately after pouring, the metal in contact with the mould walls is cooled rapidly to a temperature below the liquidus with the result that rapid columnar crystallization occurs under conditions of steep temperature gradient. In many alloy systems, particularly those with a wide freezing range, a concentration gradient may exist in the casting under these conditions. As solidification proceeds and the mould tends to become saturated with heat, the rate of heat extraction decreases. Consequently the band of freezing (i.e. the interval between the liquidus and solidus isothermals), becomes wider as the temperature gradients become shallower, until a state of dynamic heat balance is set up; thereafter the width of the band remains constant. In the state of dynamic heat balance the amount of heat evolved by crystallization is equal to that removed at the inner surface by the mould; instead of a gradual fall in temperature from the liquid centre of the casting to the solidus isothermal (or the mould wall), there is a band of freezing which proceeds towards the centre, while the temperature of the remainder of the liquid remains at or just below the liquidus temperature.

Heggie concludes that equiaxial crystallization is most likely to occur when the band of freezing is wide and the temperature gradients shallow, or when the state of

dynamic heat balance is attained. He considers that in equiaxial crystallization it is essential for "free" crystallites to form outside the range of influence of existing columnar dendrites, but he does not suggest a satisfactory mechanism by which this can occur. As regards the effect of the dynamic heat balance, as indicated by an isothermal liquidus arrest, it may be pointed out that it was found in the present investigation that in spite of the attainment of this condition, a structure consisting of columnar crystals extending right across the block from the mould walls to the centre of the casting might be produced (cf., for example, Figs. 56, 80 and 81). The absence of a central zone of equiaxial crystals from this structure is probably due to gravity settlement of the free crystallites into the lower part of the casting, permitting unrestricted columnar wall growth. This is an important factor that Heggie overlooked.

Matuschka⁽³¹⁾ attempted a correlation of the thermal conditions and macrostructures of a series of steel ingots, and suggested that the precise moment at which thermal equilibrium is established determines the transition from the columnar to the equiaxial form of crystallization. He did not attempt to explain how this transition occurs.

Although Matuschka's observations are in general agreement with the results of the present investigation of thermal conditions described on Pages 119-127, it is considered that owing to the influence of other factors,

such as the settling characteristics of independent crystallites, an exact correlation with the macrostructure cannot be made.

Northcott⁽⁷⁾ discussed the relationship between temperature and composition gradients. His suggested explanation of the formation and growth of independent nuclei ahead of the columnar mould-wall crystallization is summarized below:-

Using the results of the mathematical determinations of Lightfoot^(12c) and the evidence supplied by Matuschka⁽³¹⁾, Fig. 89 was constructed, showing the temperature distribution in a solidifying casting.

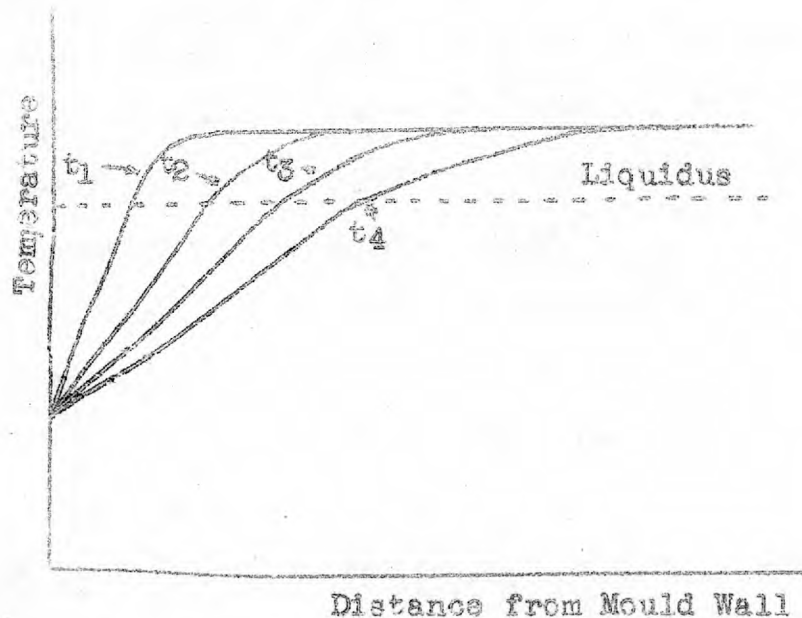


Fig. 89. Temperature Distribution in Solid and Liquid Steel at times t_1 (early) to t_4 (late)

The curves indicate that the gradients decrease with time, and throughout the central portion of the casting, there is no appreciable gradient for the greater part of the solidification time.

Generally in castings the cooling rate is too rapid to allow for the attainment of equilibrium by diffusion of the alloying elements and impurities. From a consideration of the equilibrium diagram of a typical solid solution alloy, Northcott showed how concentration gradients could arise under non-equilibrium conditions. Thus the liquid in contact with the solidification front would be highly segregated, while only a short distance away it will possess the original composition. As the liquidus temperature is determined by the composition of the liquid, it would vary, at any given time during solidification, with the distance from the mould wall; according to Northcott this variation can be represented diagrammatically by a curve such as that shown in Fig. 90. If this curve is superimposed on the approximate temperature - distribution curve (Fig. 89), the two curves may, in certain circumstances, intersect at two places. This means that conditions would then be favourable for solidification at two positions simultaneously within the casting.

Northcott considered typical sets of conditions as illustrations of the different modes of solidification.

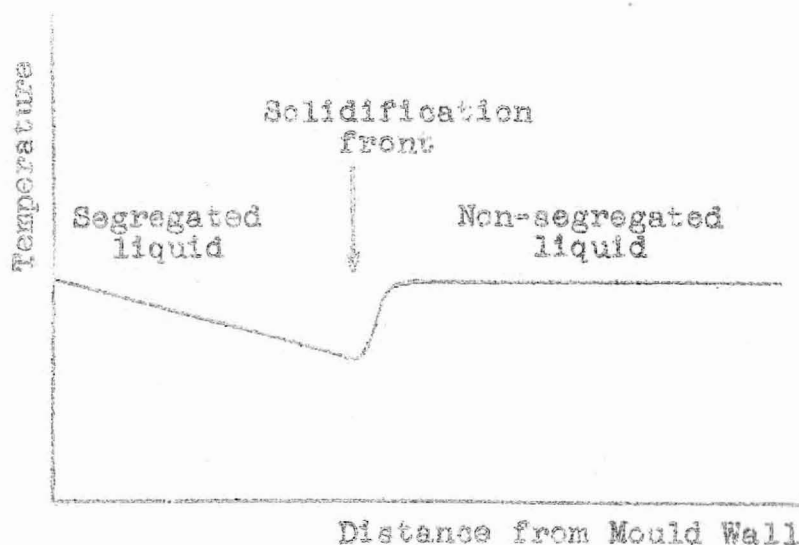


Fig. 90. Variation of liquidus temperature with distance from mould wall, at time t .

(a) High Casting Temperature, Massive Chill Mould: Under these conditions a steep temperature gradient persists throughout the course of solidification. The thermal distribution curves for given times cut the "variable liquidus curve" at one place only. This indicates that solidification occurs at one position only, in a given section, and as a result, columnar crystals grow without restriction towards the centre of the casting. (Matuschka⁽³⁴⁾ considers that under these conditions thermal equilibrium is not attained.)

(b) Normal Casting Temperature, Chill Mould: Under these conditions the temperature gradients are initially steep but flatten as solidification proceeds. During the early stages, when the temperature gradient is steep, the temperature distribution curves cut the "variable liquidus curve" at one place only, with the resulting formation of columnar crystals, as described above. However, at a later stage the temperature gradient decreases and there is an instant when the curve crosses or meets the "variable liquidus curve" at two positions simultaneously. This results in solidification at two places, one of which is independent of the solidification front which has developed from the mould wall. The presence of the "free" crystals will restrict the columnar growth, and under normal circumstances equiaxial crystals will develop in the remaining liquid.

(c) Sand Castings. Northcott considers that in the solidification of sand castings, the temperature gradient remains fairly shallow throughout the solidification process, and the temperature-distribution curve persistently cuts the "liquidus" curve at two places simultaneously. This would correspond, according to Heggie and Matuschka, to the rapid attainment of thermal equilibrium conditions, with the result that independent nuclei would soon form and grow in the manner outlined in the preceding paragraph. According to Northcott, the solidification of the central mass of liquid is associated with the establishment of a series of concentration

gradients. During this time the casting is in a "mushy" state, due to the formation and development of equiaxial crystals in this central area.

It is important to note, however, that more recent temperature measurements in solidifying castings made by Bishop, Brandt and Pellini^(63,64), and also in the present investigation, have indicated that the central mass of liquid cools relatively rapidly to the "liquidus" temperature; thereafter the temperature gradient in this liquid is negligible. In the light of these results it is necessary to modify the theory proposed by Northcott, since his temperature curves show the temperature of the central mass of liquid to be appreciably above the liquidus. The necessary modifications however make no essential difference to the theory.

Although Bishop, Brandt and Pellini^(63,64) have not considered the columnar/equiaxial transition in their investigations, the results of their temperature-measurement experiments on sand and chill castings indicate two different modes of solidification, which are presumably associated with the different modes of crystallization. They consider solidification to be basically a problem of heat transfer, and have studied the methods by which the total heat content of the liquid metal is extracted by different mould materials.

In the case of solidification from a chill mould, there was a high rate of heat extraction, and steep temperature

gradients were established throughout the solidifying metal. Under these conditions a narrow band of freezing travelled from the surface of the casting to the centre. (The narrow freezing band is a direct result of the close proximity of the liquidus and solidus isothermals). With this type of solidification, it was considered probable that the rate of extraction of the superheat was insufficient to satisfy the chilling effect of the chill mould. Consequently a considerable amount of latent heat was recovered from the metal in contact with the mould wall. Under these conditions the state of dynamic balance was not attained.

However, in the case of solidification from sand walls, the rate of heat extraction was much lower. Consequently the temperature gradients induced in the solidifying metal were much shallower and the final solidification time greater. Under these conditions it was found that although solidification proceeded progressively across the block from the wall towards the centre, crystallization began at the centre of the casting before solidification had been completed at the mould wall. The rate of heat extraction by the mould was so small that the mould/metal interface temperature was maintained above the solidus temperature by the evolution of latent heat from the interior of the casting.

Temperature determinations obtained during the solidification of representative sand and chill castings illustrate these various features.

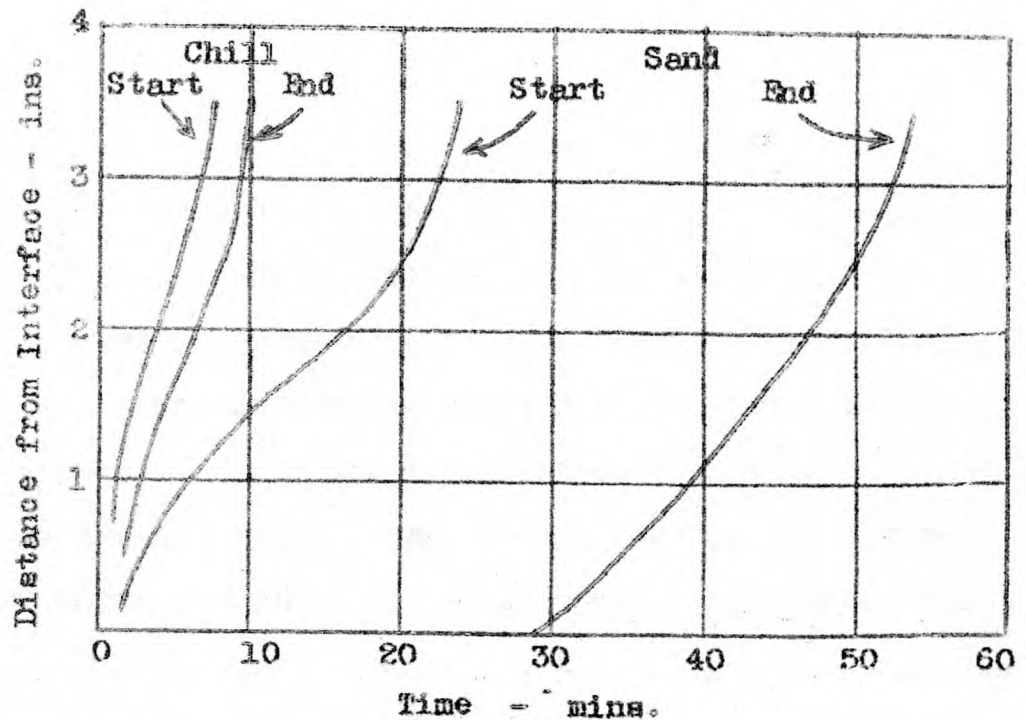


Fig. 91. Comparison of Solidification Curves for chill and green sand castings

(b) Consideration of the Results of the Temperature

Measurements obtained during the Present Investigation
in relation to the Solidification Process.

By considering the temperature distribution in solidifying castings, as determined by direct measurements in the present investigations, it is possible to picture the mode of solidification and crystallization of vertical castings, and to see how this accords with the views suggested by Heggie, Northcott and Bishop, Brandt & Pellini.

Let us consider the vertical top-fed and bottom-fed blocks, the thermal investigation of which has been described in pages 109 - 129.

Top-fed Block. Cooling curves for various positions

across this type of block were shown in Figs. 77 and 78. From these curves the upper series of diagrams in Figs. 92 and 93 have been drawn. They represent the temperature gradients existing in the solidifying castings at different times after pouring until the completion of solidification.

The middle series of diagrams (Fig.92) and the lower ones in Fig.93 show the percentage of solid existing at any given distance from the side walls, at different times after pouring. These values were calculated from the equilibrium diagram (Fig.76), and the temperature gradients.

The bottom diagrams in Fig.92 represent the results of the "pour-out" experiments and show the approximate extent of solidification after one, two, three and four minutes.

These diagrams, interpreted in conjunction with the appropriate cooling curves, suggest the manner in which the solidification of casting proceeded.

The absence of a liquidus arrest in the cooling curve recorded by the couple placed close to the mould wall, (Fig.77, p.120 - Curve 1), indicates that the metal in contact with the mould wall solidified rapidly after pouring. Moreover, the temperature fell to the liquidus value throughout the casting in less than a minute after pouring. Consideration of the top and bottom diagrams in Fig.92(A) shows that the columnar wall growth developed under conditions of steep temperature gradient within the band of freezing,

viz. 140°C per inch. After one minute (Fig.9 A) the columnar crystals extended about 0.5" into the liquid. The metal had completely solidified to a depth of 0.05" from the mould wall, whilst in the plane of the tips of the dendrites the proportion of liquid approached 100% (as indicated by the middle diagram).

These results show that the apparently solid wall indicated by "pour-out" tests, actually consisted of varying percentages of liquid and solid, there being in fact only about 5% solid at the inner surface of the "solid" wall. On up-ending the casting, the liquid had remained entrapped between the dendritic arms and branches.

After two minutes cooling, solidification had proceeded to a distance of 0.8" from the mould wall as indicated by the top and middle diagram in Fig.92B. The temperature gradient was less steep during this period, though still sufficient to produce columnar crystals; the gradient shown in the diagram corresponds to approximately $68^{\circ}\text{C}/\text{in}$. Judging by the section of the bled casting (Fig.84 or 92B), it might be imagined that solidification had progressed to completion to a depth of 0.6", but in fact the completely solidified shell of metal extended to only about 0.075" from the mould wall, and as comparison of the middle and lower diagrams shows, all the material over 5% solid was retained at the walls when the casting was up-ended.

During the third minute solidification proceeded

with the formation of small nuclear or equiaxial crystals in place of the columnar form, under conditions of relatively shallow temperature gradients; the mean value of the gradient between X and Y, Fig.92C, top diagram, is $19^{\circ}\text{C}/\text{in}$. At the end of this period, solidification had progressed to a depth of 1.00", the completely solid shell of metal extending for a distance of 0.125" from the mould wall. Upon up-ending the casting, all material more than 15% solid adhered to the solidifying wall. The increased percentage of solid in the metal bled away was probably due to the comparatively loose packing of the nuclear/equiaxial crystals. The inner surface of the bled casting (Fig.85), had a rough appearance, indicative of the loosely-packed nuclear/equiaxial type of growth, and the thickness of the "solid" shell varied considerably, possibly as the result of the movement of small crystals towards the head during the up-ending process.

After four minutes, solidification was proceeding across the entire section of the block, the temperature at the centre of the block having fallen 1°C below the liquidus value. (Fig.92D). Moreover, Fig.79 (p.122) shows that after four minutes the liquidus arrest had been completed along the length of the block, and hence solidification was proceeding throughout the whole casting and not only across the horizontal section to which the upper diagrams in Fig.92 refer. The upper diagram, figure 92D, suggests that during the previous minute the temperature gradient in the central

portion of the block had been comparatively shallow; the resulting crystal structure was of the nuclear/equiaxial form. The completely solid shell of metal extended to a depth of 0.2" from the mould wall. Up-ending of the casting under these conditions resulted in the movement of a portion of the partially solid mass towards the head, leaving cavities in the central portion of the block (Fig.86).

Between the fourth and seventh minutes after pouring, further cooling resulted in the progressive completion of solidification from near the mould wall to the centre of the block. Figs 93, E,F,G,H show that the mean temperature gradient under which the central nuclear/equiaxial crystals developed increased from $24^{\circ}\text{C}/"$ after five minutes to $37^{\circ}\text{C}/"$ after six and a half minutes.

Another interesting feature revealed by Fig.93 (E,F,G,H) is the final rapid solidification at the centre of the block. This is clearly shown by the data tabulated below.

Time After Pouring.	Depth of Completely Solid Metal	% Solid at centre of block.
5 minutes	0.35"	35%
6 "	0.62"	43%
$6\frac{1}{2}$ "	0.90"	57%
7 "	1.25"	100%

The cooling curve obtained from the readings of a thermocouple placed at a position near the head (Fig.79 p.123, Curve 7), shows that the solidus temperature and hence final solidification of the block, was reached 8 minutes 48 seconds after pouring.

Let us consider now the evidence provided by the differential temperature measurements. The temperature difference curves obtained between positions on the transverse section of the top-fed vertical block (Fig.74, p.114), indicate that although initially conditions were turbulent within the interior of the casting, there was a rapid removal of heat from the metal in contact with the mould wall, followed later by a series of increases in temperature difference between successive positions as solidification proceeded across the block. These results accord with the mode of solidification indicated by the direct temperature measurements (Fig.77) and are discussed in detail below.

A consideration of the direct temperature measurements (Fig.77) shows that during the first half minute or so the temperature of the liquid fell to the liquidus value. However, during this time the temperature indicated by the thermocouple placed close to the mould wall (Fig.77, position 1 and Fig.74, position d) fell rapidly; on the other hand that of position 2 (Fig.77) or C (Fig.74) remained at the liquidus value for almost one minute. Consequently during that minute the temperature difference between positions "c"

and "d" (Fig. 74) continued to increase. After approximately $1\frac{1}{2}$ minutes, solidification had commenced at position "c" (as is indicated by the end of the liquidus arrest (Curve 2, Fig. 77) and as a result the temperature difference between "c" and "b" increased as position "c" cooled and solidification proceeded towards "b"; this is indicated by the rise of curve 2 (Fig. 74) for about $2\frac{1}{2}$ minutes after pouring. However, as the solidification front proceeded between "c" and "b" the temperature of both "c" and "d" continued to fall. (See direct curves 1 and 2, Fig. 77). If the temperature at each of these positions fell at a uniform rate the difference curve would be horizontal, but if the position "d" cooled faster than "c" the temperature difference curve would continue to rise. (If "d" cooled slower than "c" the curve would fall.) Curve 3 (Fig. 74) slowly rose after solidification had reached position "c" then remained constant for a while and finally fell. This indicates that position "d" cooled slightly faster than "c" up to a time of $6\frac{1}{2}$ minutes after pouring, then at the same rate until about 8 minutes and finally slower.

Similarly the temperature difference curves show that as solidification was proceeding between positions "b" and "a" position "c" continued to cool slightly faster than position "b" until about 7 minutes after pouring. Thereafter it cooled at a slower rate even though the temperature of "c" was lower than that of "b".

When the solidification front arrived at the centre

of the block ("a") about 4 minutes after pouring, curve 1 (Fig.74) indicates that the temperature-difference between "b" and "a" was constant for about two minutes. It then increased to a maximum after $7\frac{1}{2}$ minutes and finally decreased.

It is of interest to note that during the time that the solidification front was proceeding between position "c" and "b", the direct temperature curves (Fig.77) indicate that positions "a" and "b" were at the liquidus temperature. The difference curves, however, indicate the existence of a temperature difference of 2° or 3° C between these points. This is considered to be due to experimental error, which could not be eliminated.

Comparison between the direct and difference curves suggests that the change in the slope of the difference curves after approximately $7\frac{1}{2}$ minutes corresponds either with the completion of solidification after 7 minutes, or the peritectic transformation after $6\frac{3}{4}$ minutes. Since the peritectic has a marked effect on the slope of the direct curves, whilst the solidus has none, it is considered that the discontinuity recorded in the difference curves is the peritectic transformation.

The temperature-difference curves obtained for positions along the central axis of the top-fed block (Fig.72, p.110) are basically similar to those obtained from the transverse positions. By considering them in relation to the direct curves, it is possible to trace solidification along

the length of the block, beginning at the end wall and concluding at the head.

Due to turbulence, and the rapid fall in temperature of the interior of the casting to the liquidus temperature, there was an initial period of variable temperature difference. This lasted for about two minutes in the upper half of the block and for about one minute in the lower. However, half a minute after pouring, curve 1 (Fig. 72) indicates that as solidification proceeded from the bottom wall the temperature of position "g" fell rapidly, while that of position "f" remained at the liquidus temperature. After $2\frac{1}{2}$ - 3 minutes, curve 2 shows that a temperature difference arose between "f" and "e", thereby indicating that solidification was proceeding between "f" and "e". While this was occurring, position "g" continued to cool faster than "f" and as a result curve 1 continued to rise. Finally after 4 minutes solidification had progressed to position "e" and the temperature difference between "g" and "f" attained a maximum, after which "f" cooled faster than "g", although it remained hotter than "g".

The process of solidification continued in a similar manner along the axis of the block towards the head. After about 8 minutes when solidification (or the peritectic arrest) was almost complete in all parts of the block, there existed a gradually increasing temperature gradient from the head to the bottom of the block. This steadily decreased as

cooling proceeded to room temperature.

During the solidification process the rate of extraction of heat by the sand mould decreased. This is reflected in the difference curves by the decrease in the maximum value of the temperature-differences between each set of positions, as solidification proceeded along the length of the block.

As was mentioned on page 111, in curves 4, 5 and 6 (Fig.72), the temperature differences decreased before finally increasing to a maximum. The reason for this is probably that solidification was occurring throughout the central portion of the block (due to solidification from the side walls). That is to say, solidification was occurring ahead of the main solidification front from the bottom wall, and this resulted in a temporary temperature reversal as indicated by a dip in the curves. The fact that crystallization was occurring throughout the interior of the block relatively early in the process, well ahead of the arrival of the main solidification front from the bottom, is illustrated in Fig.79, p.122, by the general completion of the liquidus arrest at all levels above that corresponding to curve 2, within about 4 minutes. Nevertheless, final solidification proceeded progressively from the bottom of the block towards the head.

The structures of the bled castings help to illustrate the manner of solidification along the length of

the block, from the end wall towards the head. (Figs. 83-86, 119). It might be expected that the depth of solidified metal would be decidedly greater from the bottom end than from the side walls, because of "settling" action of the crystallites and more rapid removal of heat remote from the head. However, the results show that this is not the case, the difference being quite small. See Fig. 92 A, B and C. It will be seen, however, by comparing section B with section A, that the form of crystallization had changed after about one minute to a nuclear/equiaxial type, the impression given by the structure being that crystallites had settled toward the bottom of the casting, preventing further columnar growth from the bottom wall.

Bottom Fed Block.

The process of solidification of this type of casting is illustrated by Figs. 94 and 95 in which the upper diagrams show the temperature gradients existing horizontal across the block, midway between the top wall and the feeder, at a series of time intervals after pouring. These diagrams have been derived from the series of cooling curves in Figs. 80 and 81, and the lower diagrams have been derived from those above by means of the equilibrium diagram.

It is evident from curve 1 (Fig. 80) that the molten metal in contact with the side wall must have begun to solidify very soon after the mould was filled: at position 1, 0.13" from the wall, the temperature had fallen below the

liquidus value in less than half a minute. Within a minute, all the liquid right through to the centre of the block, had cooled to the liquidus temperature. It was not practicable to bleed the bottom-fed castings, but the macrostructure of a similar block, cast under the same conditions (Fig.56) shows that columnar crystals grew horizontally from the side walls, meeting in the centre of the block.

Reference to Fig.94A shows that after one minute the tips of the columnar crystals extended about 0.5" into the liquid. A moderately steep temperature gradient existed within the band of freezing, of the order of $55^{\circ}/"$. Extrapolation of the temperature gradient curves from the position of the couple nearest the mould wall to the metal-mould interface suggests an interface temperature of approximately 1478°C . This temperature is above the solidus value of the steel (1464°C), which means that the metal in contact with the mould was only 75% solid (lower diagram Fig.94A).

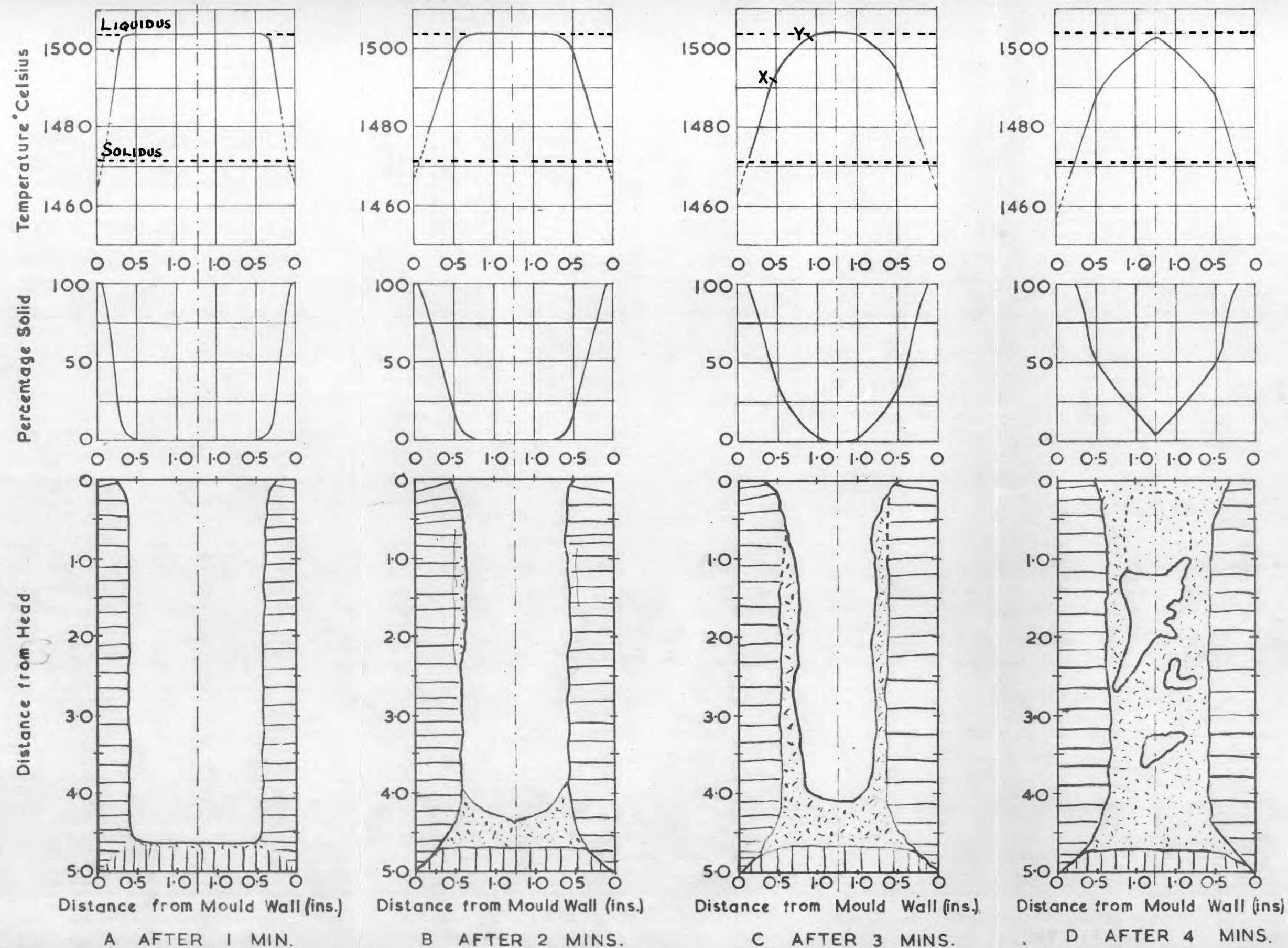
After two minutes cooling, the columnar crystals extended to a distance of 0.75" from the mould wall (Fig.94B). The temperature gradient remained approximately the same as for the first minute, viz. $50^{\circ}/"$. Extrapolation of the temperature-distribution curve suggests a metal/mould interface temperature of 1470°C , which is still above the solidus value and corresponds to 92% solid (lower diagram, Fig.94B).

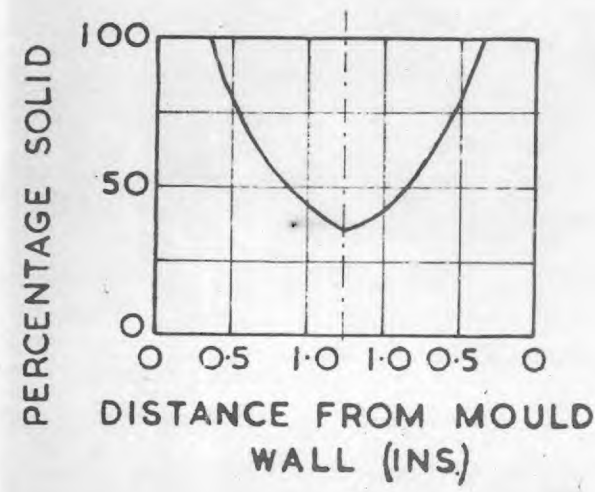
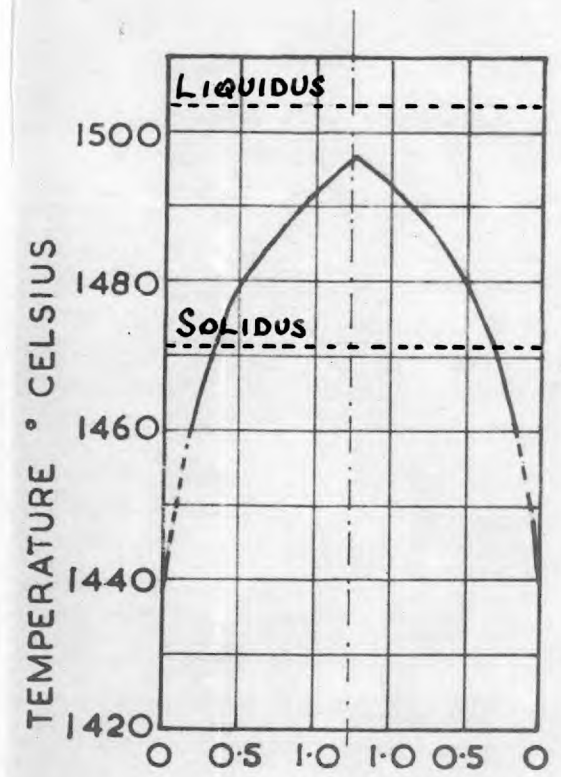
At the completion of the third minute, the columnar crystals reached nearly to the centre of the block, and the

metal at the interface was completely solid. The temperature gradient in the band of freezing near the mould wall was almost as steep as before, viz. $46^{\circ}\text{C}/\text{in}$, but towards the centre it was considerably shallower. The cooling curves in Fig.80 and 81 indicate that the columnar crystals reached the centre of the block about $3\frac{1}{2}$ minutes after pouring; but there were still some regions completely liquid lower down in the block, near the feeder, as is shown by the curves in Fig.82.

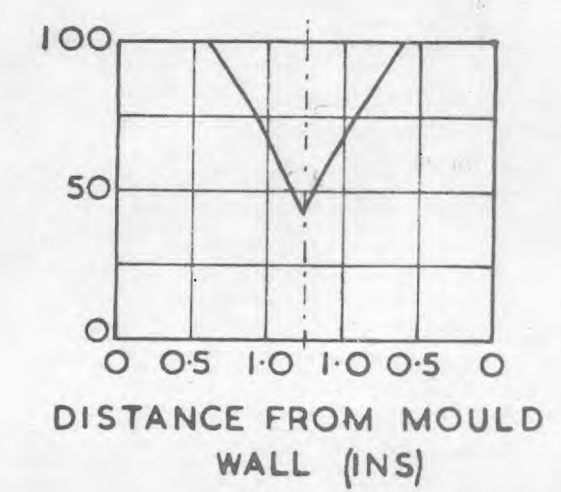
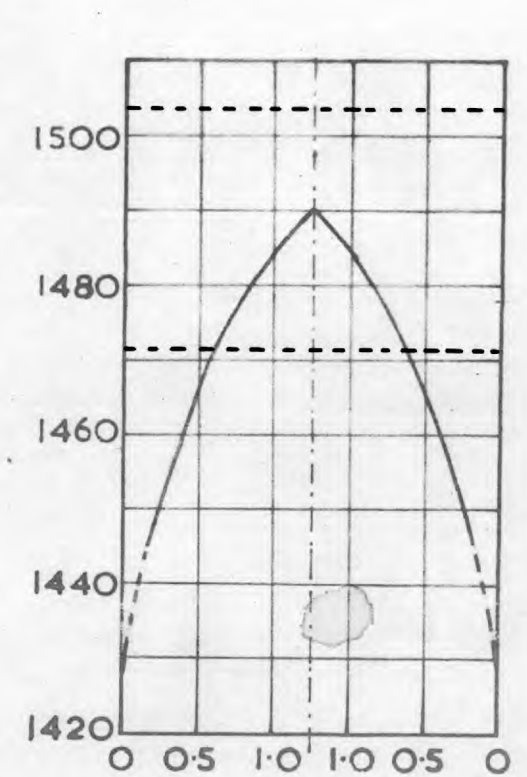
After four minutes (Fig.94D) the completely solid shell extended to about 0.2" from the mould wall, and the metal at the centre was 15% solid. Figure 82 p.126 indicates that solidification was now proceeding throughout the casting, right down to the bottom of the block.

During the next two minutes, solidification proceeded to completion at a rate that steadily increased because of the rapidly decreasing volume of liquid remaining at the centre of the casting. (~~This increase in the rate is further discussed on Page~~). Figure 95 shows that at the end of five minutes the completely solid shell extended 0.5" from the mould wall. The metal at the centre was 35% solid. Similarly, figure 95F shows that after a further half minute the solid shell was 0.7" thick and the centre 50% solid. Although this section of the block was completely solid at the end of the 6th minute, Fig.82 shows that the solidification of the lower portion of the block was not completed until 7 minutes 24 seconds after pouring.

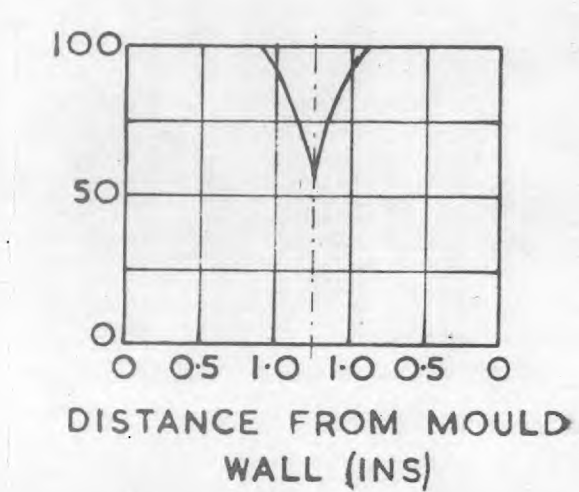
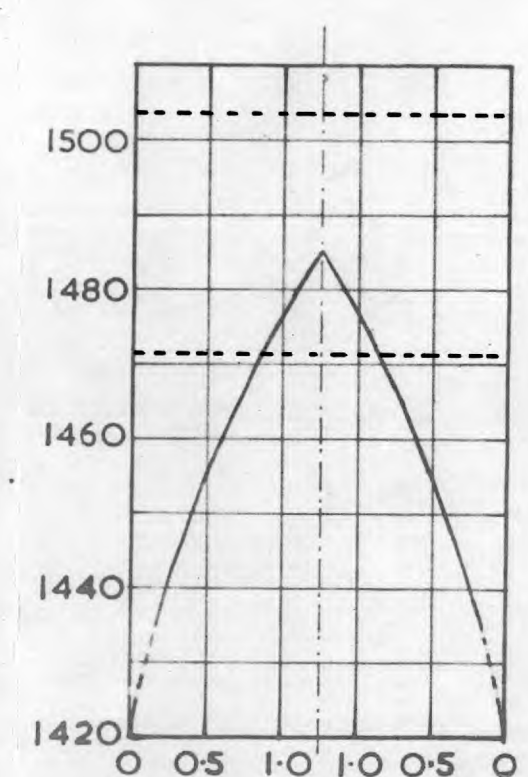




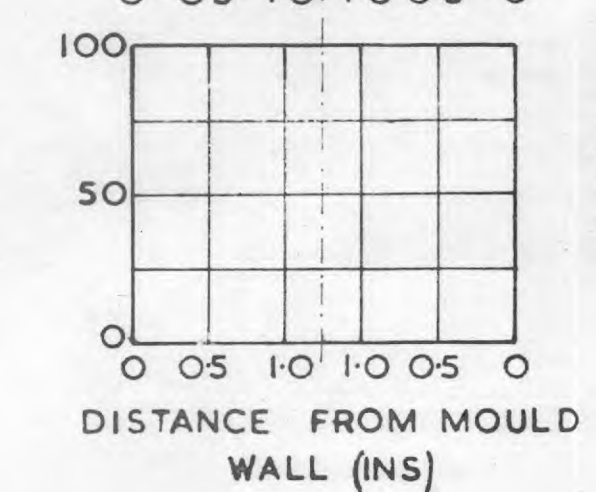
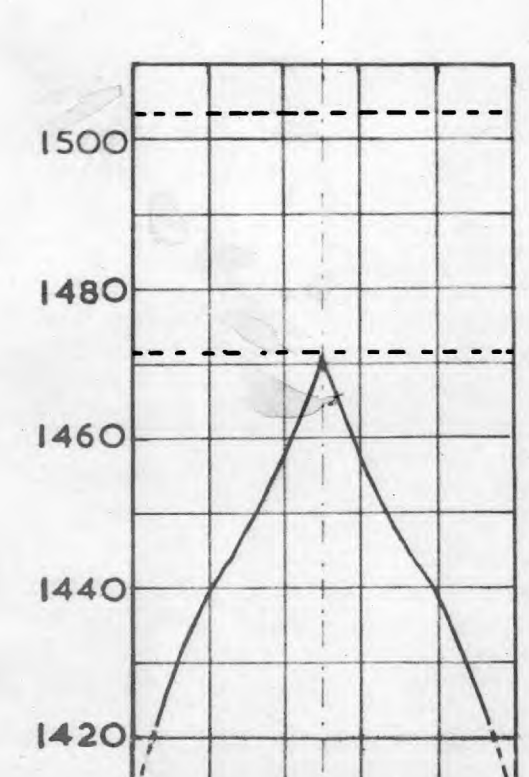
E AFTER 5 MINS.



F AFTER 6 MINS.



G AFTER 6.5 MINS.

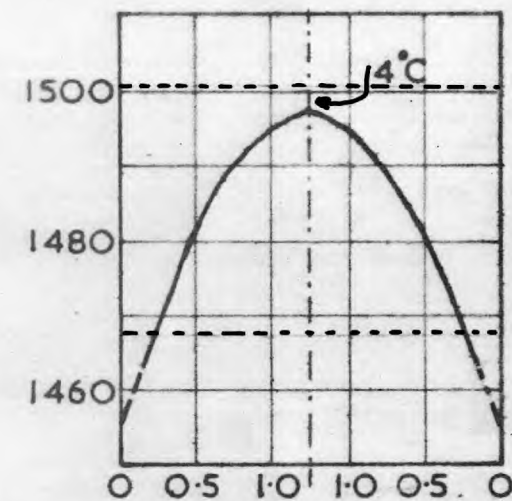
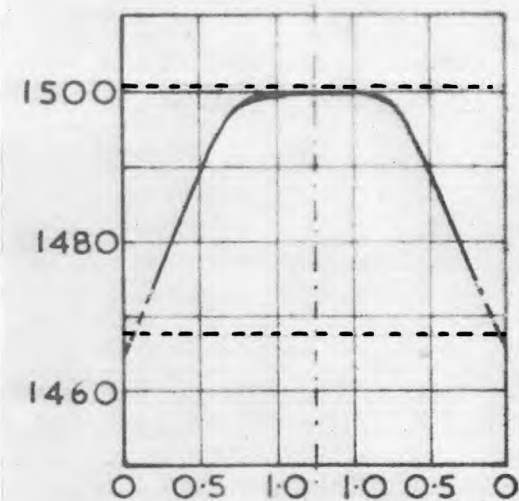
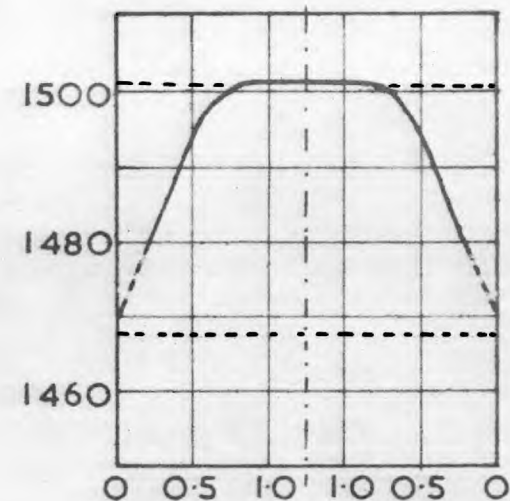
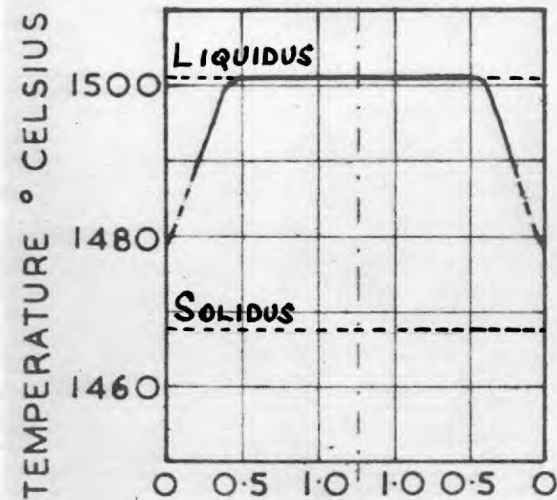


H AFTER 7 MINS.

TEMPERATURE GRADIENTS IN SOLIDIFYING CASTING.

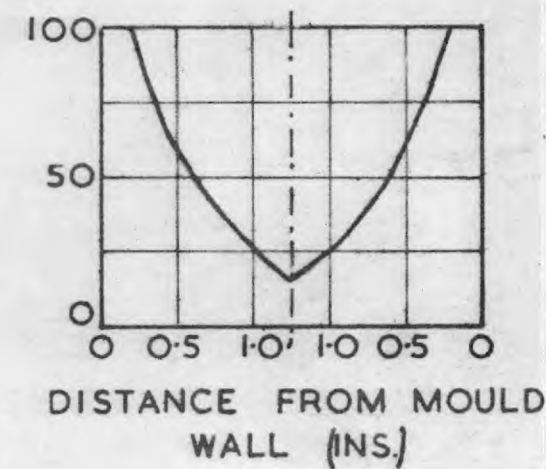
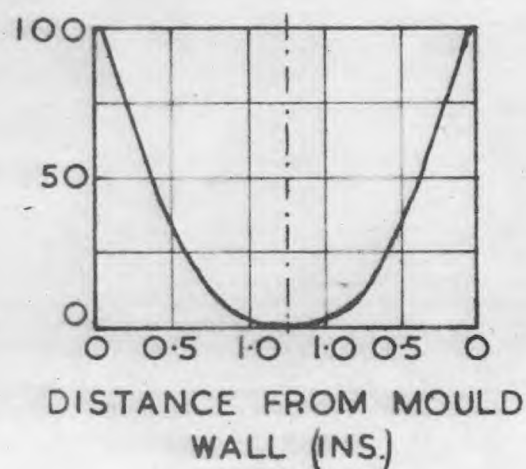
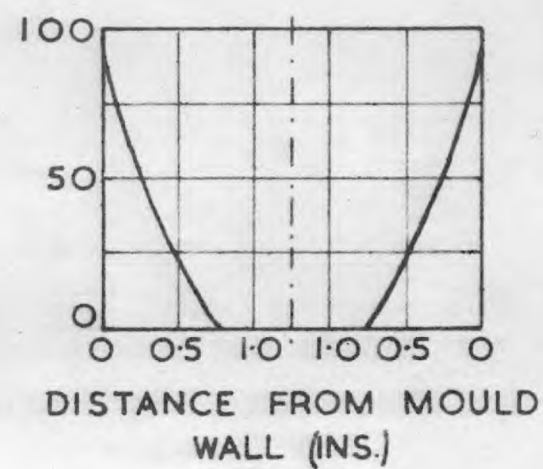
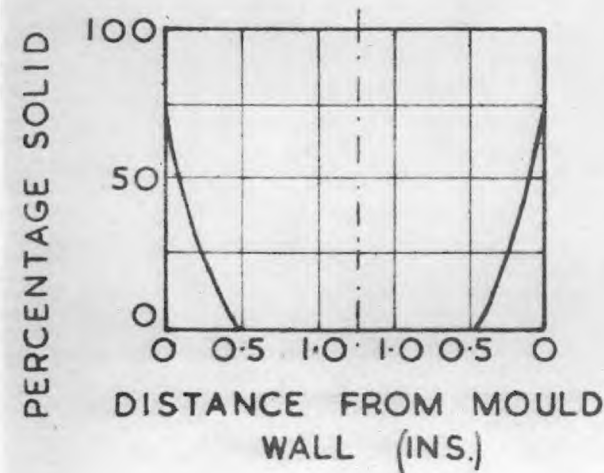
FIG.93
TOP-FED CASTING
TYPE 'A'

PERCENTAGE SOLID IN SOLIDIFYING CASTING.



TEMPERATURE
GRADIENTS IN
SOLIDIFYING
CASTING.

FIG.94
BOTTOM - FED
CASTING, TYPE 'E'



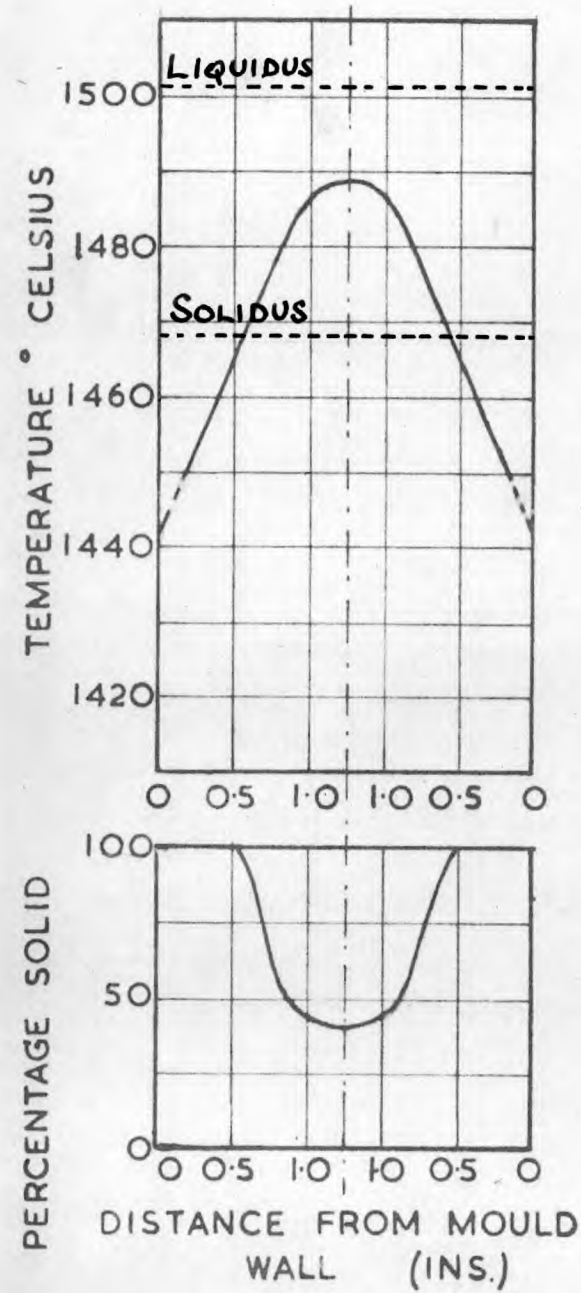
PERCENTAGE SOLID
IN SOLIDIFYING
CASTING.

A AFTER 1 MIN.

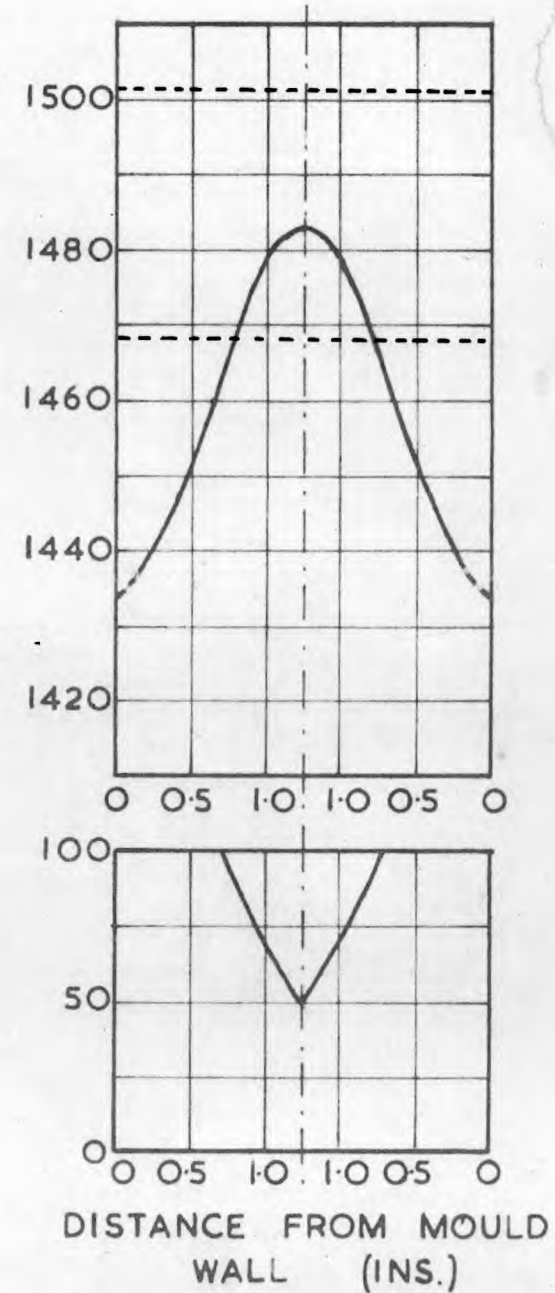
B AFTER 2 MINS.

C AFTER 3 MINS.

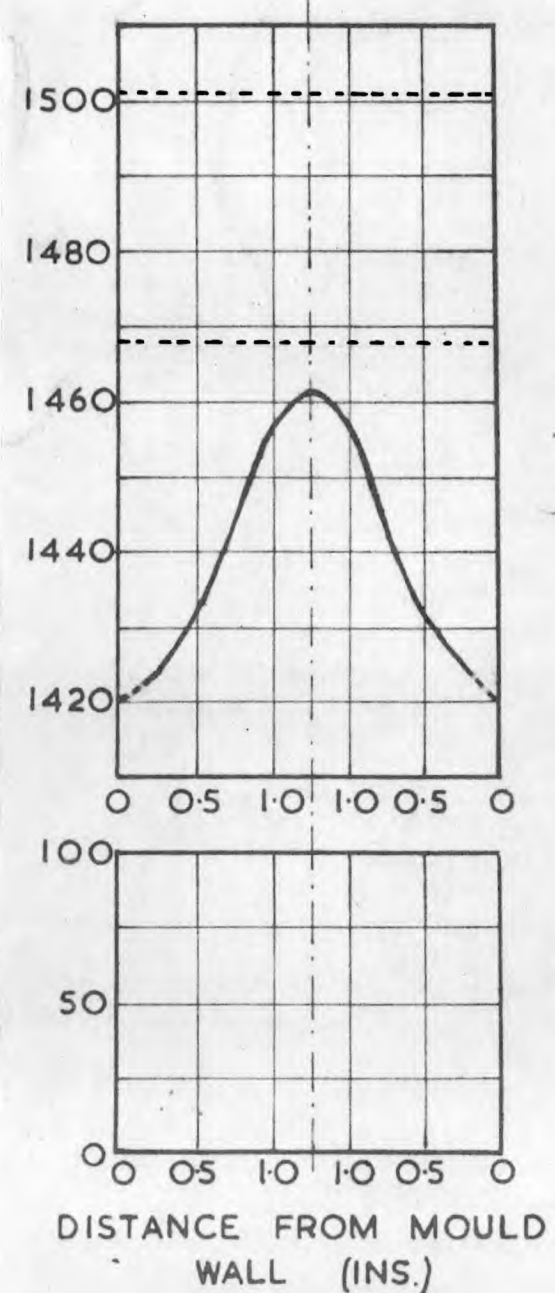
D AFTER 4 MINS.



E AFTER 5 MINS.



F AFTER 55 MINS.



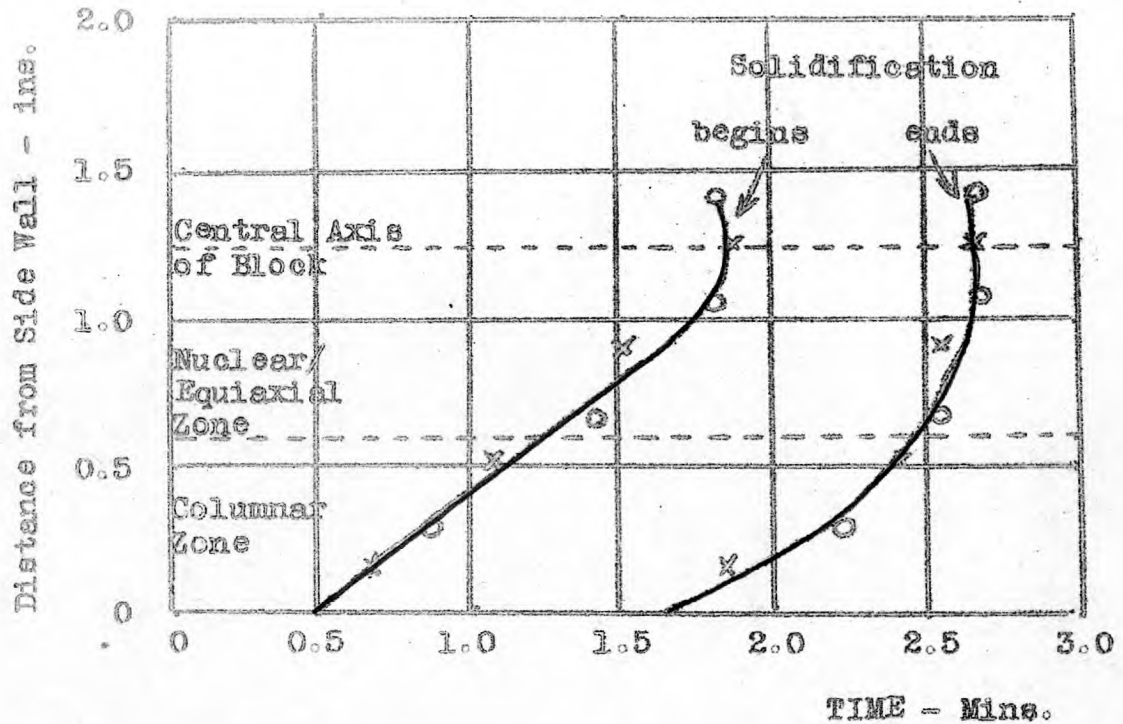
G AFTER 6 MINS.

TEMPERATURE
GRADIENTS IN
SOLIDIFYING
CASTING.

BOTTOM-FED
CASTING, TYPE 'E'

PERCENTAGE SOLID
IN SOLIDIFYING
CASTING.

FIG. 96.



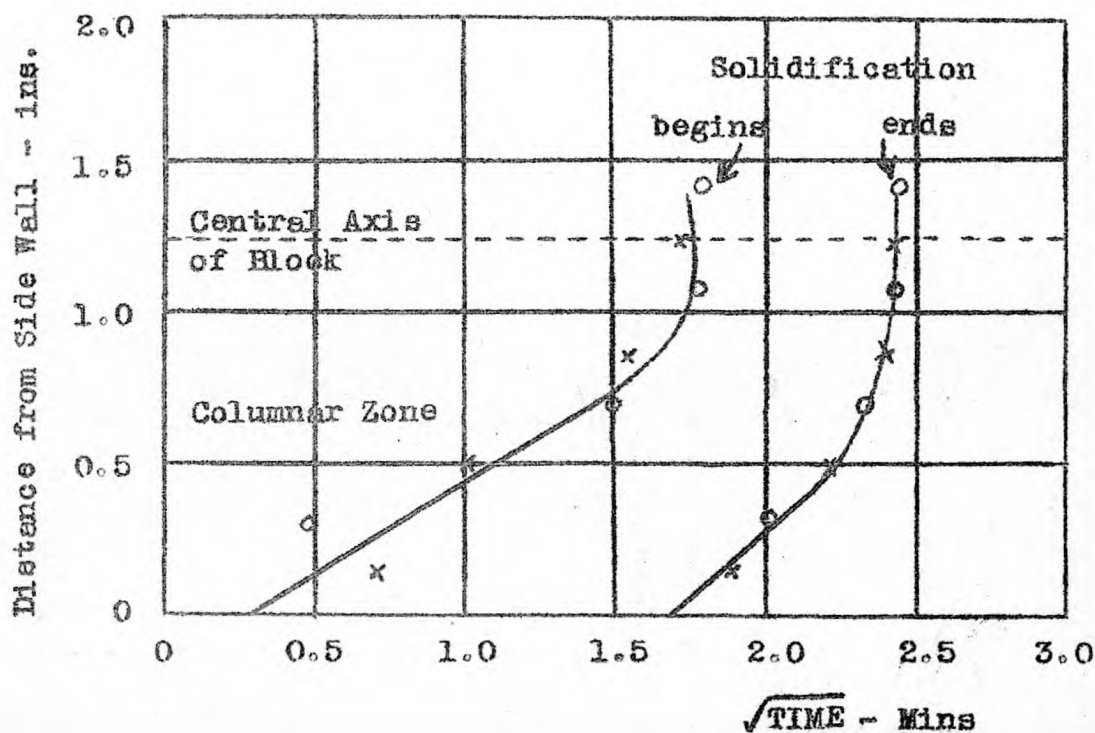
Thermocouple		1a	1b	2a	2b	3a	3b	4a	4b
Distance from Mould Wall (inches)	D	0.18	0.30	0.53	0.63	0.89	1.06	1.25	1.42
End of Liquidus Arrest (Minutes)	T	0.44	0.75	1.15	2.00	2.25	3.25	3.50	3.25
	$\sqrt{T}$	0.66	0.87	1.07	1.41	1.52	1.80	1.82	4.80
Solidus (Minutes)	T	3.45	3.20	5.83	6.45	6.95	7.10	7.00	7.10
	$\sqrt{T}$	1.86	2.24	2.41	2.54	2.54	2.67	2.65	2.67

Fig. 96. Solidification Curves of Castings Al and A2.

(Al:- C 0.24; Si 0.08; Mn 0.24 Superheat 74°C

(A2:- C 0.29; Si 0.10; Mn 0.43 Superheat 75°C

FIG.97.



Thermocouples		1a	1b	2a	2b	3a	3b	4a	4b
Distance from Mould Wall (inches)	D	0.13	0.32	0.50	0.70	0.86	1.08	1.23	1.43
End of Liquidus Arrest (Minutes)	T	0.50	0.25	1.00	2.25	2.40	3.20	3.00	3.25
	$\sqrt{T}$	0.71	0.50	1.00	1.50	1.55	1.80	1.73	1.80
Solidus (Minutes)	T	3.60	3.45	4.45	5.45	5.70	5.45	5.45	5.90
	$\sqrt{T}$	1.90	2.00	2.21	2.34	2.40	2.44	2.44	2.43

Fig.97. Solidification Curves of Castings E1 and E2.

E1:- C 0.28; Si 0.09; Mn 0.34 Superheat 74°C

E2:- C 0.27; Si 0.12; Mn 0.49 Superheat 70°C

West⁽¹⁶⁾, in an investigation of the thermal distribution within horizontal steel blocks cast under different conditions of superheat, composition and chilling, found that when the depth at which the liquidus arrest was completed was plotted against the square root of the time, the points lay in two approximately straight lines, representing solidification from the top and bottom surfaces respectively. These lines agreed well with the results of "pour-out" experiments made on similar blocks cast under the same conditions. It was concluded from this that the results of "pour-out" experiments represent a state of solidification corresponding to the formation of a close network of dendritic solid. If the results of the present thermal determinations are plotted in a similar manner (Figs. 96 and 97), the points also lie in a straight line for the early stages of solidification (from the side wall of a vertical block). In the later stages, however, as solidification proceeds towards the centre, there is marked deviation from the linear relationship, as the rate of solidification increases. The solidification constants k and c as determined from the slope of the straight line portion of the curves are also similar to those found by West.

A plot of the depth of complete solidification (i.e. the depth of the solidus isothermal temperature) against the

square root of the time shows little evidence of a straight line relationship, even during the earlier stages of solidification. In the later stages the rate of solidification is considerably increased as the remaining volume of liquid in the central portion decreases.

The curves indicate that there exists a wide band of freezing during the solidification of sand castings. This is in agreement with the results obtained by Bishop, Brandt and Pellini, (described on Page 153). Furthermore, the width of the freezing band for the top-fed block (Fig 96) is wider than that for the bottom-fed block (Fig 97) in the later stages of solidification. This is consistent with the view that nuclear/equiaxial crystals tend to form when the band of freezing is wide and the temperature gradient shallow, whereas a narrower band of freezing and steeper gradient favours columnar growth.

(c) "Pour-Out" Experiments.

The results of the present investigation have confirmed the view that the "solid" shell seen in a section of a bled casting actually consisted largely, at the time when the casting was bled, of a network of dendritic axes and branches, enclosing a considerable amount of interdendritic liquid, which was retained by the network during the pouring-out and which solidified in situ. In the case of columnar growth, the amount of liquid retained within the crystal axes

and network at the time of "pour-out" can be as much as 95% (P.157) at the inner surface of the shell, decreasing towards the mould wall and reaching zero at the solidus isothermal, whereas in the looser nuclear/equiaxial type of formation the maximum amount of liquid retained is lower, about 85% (P.158).

If a casting having a wide solidification front, associated with central equiaxial/nuclear crystallizations is bled, little liquid is removed, the main effect being the movement of the partly solidified steel towards the draining exit (Figs. 86, 92D). This fact has been noted by previous workers, including Nelson⁽¹⁷⁾ and Schwartz⁽²⁴⁾. They found a quantity of "mushy" material inside castings which had been inverted in the later stages of solidification.

It may be concluded therefore, that the "pour-out" method for determining the progress of solidification is liable to give misleading results, particularly in the later stages of solidification.

Interface Temperature and Superheat.

As was pointed out on page 166, an estimate of the approximate value of the sand/metal interface temperature can be obtained by extrapolating the temperature-gradient curves in Figs. 92-95. For the top-fed vertical casting, one minute after pouring, it is approximately 1465°C and for the bottom-fed vertical casting 1478°C . These temperatures are high, especially the latter, which is actually above the solidus temperature of the steel. As it appears that the superheat

and interface temperature of a casting are closely related, and that both have an important effect on the mode of solidification and crystallization, it is necessary to consider this question in some detail.

(1) Interface Temperature.

Interface temperatures quoted in the literature show considerable variance. Briggs and Gezelius⁽¹⁹⁾ give a value of approximately 1300°C ; Chvorinov⁽²⁰⁾ 1458°C for a 0.24% carbon steel. West⁽¹⁶⁾ reported a temperature of 1413°C . This was recorded by a thermocouple enclosed in thin silica tubing and placed in contact with the mould wall. By extrapolation of readings obtained from a series of thermocouples placed in the solidifying metal, he estimated that the interface temperature, less than 1 minute after pouring, was 1482°C for a 0.47% carbon steel cast with a superheat of 127°C . Beeley⁽⁷⁰⁾ has also shown that the interface temperature is dependent on the casting temperature and quotes a figure of 1442°C for a casting temperature of 1575°C .

From the above evidence it appears that whilst a temperature of 1465°C may be reasonably expected, the value of 1478°C obtained for the bottom fed block is somewhat high. However, if the manner in which the metal enters the mould is considered, it seems possible that 1478°C may not in fact be too high. During pouring, the metal entering the bottom-fed block rises gradually from the base to the highest point, heating the mould walls as it passes over them. Fresh hot

metal enters and forces the cooler metal towards the top. Thus at the end of pouring, the mould walls are considerably preheated by the passage of hot metal over the surface. In the case of the top-fed block, the metal immediately fills the bottom of the mould, fresh metal entering from the head until the mould is full, without effectively displacing the chilled liquid from the lower part of the block. Under these circumstances there is no opportunity for the mould wall to be preheated, and one would expect the interface temperature to be lower.

As a result of the preheating effect of the passage of liquid metal (superheated about 70°C) over the mould walls in the bottom-fed castings, the rate of heat extraction is sufficiently low to allow the interface temperature to remain for some time above the liquidus value. Thus the final solidification at the surface does not occur until the solidification front has progressed appreciably towards the centre of the casting. In the case of the solidification of the top-fed castings, the mould walls are not "pre-heated" and the superheat is insufficient to maintain the interface temperature above the solidus. On the basis of Matuschka's⁽³¹⁾ explanation of solidification, the latent heat of solidification is extracted so rapidly that the state of dynamic heat equilibrium is not attained. Consequently, a narrow band of freezing proceeds across the casting from the mould wall towards the centre.

However, although the initial rate of heat extraction is probably greater in the top-fed blocks than in the bottom-fed type, it was found that the columnar crystallization (normally associated with high rates of heat extraction and steep temperature gradients) extended to the centre of the bottom-fed casting and rather less than half way to the centre in the case of the top-fed casting. Furthermore, the latter had a longer solidification time than the former. This apparent anomaly is difficult to explain fully, although it is probably due to two effects - (a) "settling" of crystallites and (b) the nature of the heat transfer in the system. Unhindered settlement of the crystallites into the feeder of the bottom-fed casting (discussed more fully on page 177) apparently enabled the columnar crystallization to develop unrestricted by nuclei present in the interior of the casting. In the top-fed type, however, the crystallites would pile up on the bottom, and would also probably be fed continuously into the block from the head. Such an explanation does not satisfactorily explain the different times of solidification, which are governed by rates of extraction of heat. A comparison of the mould designs, however, shows that although the weight of metal employed in each casting was approximately the same, the surface area of the bottom casting was much greater. Consequently, while the latter type of casting was being poured, a greater proportion of the superheat would be absorbed by the mould walls. Support

for this view is provided by the fact that it was impossible to keep the head open for a sufficient time to feed the block efficiently (Castings WX3, WX1, WOX5, Fig.56) without employing exothermic cores to prolong the solidification of the feeder head. Thus, although the bottom method of feeding produced "preheated" moulds the final rate of solidification was faster than that of the top-fed block.

(ii) Superheat.

The effect of superheat in relation to nucleation and growth within a solidifying casting was discussed in the earlier section (Page 132). It was concluded that one important effect of superheating was to eliminate some of the "potential" nuclei in the liquid, with the result that when the metal was cast a coarse crystal structure developed from the remaining nuclei. Consideration of the effect of superheat on the columnar/equiaxial transition was deferred until the basic mechanism had been described.

Refs.

There is experimental evidence (16,63,64) to support the view that superheating prolongs the solidification time of a casting. This is not necessarily the result of a decrease in the rate of solidification, but is due rather to an increase in the initial lag before solidification starts at the surface. The initial lag is the time necessary for the removal of the superheat (i.e. the excess heat present in the system) from the liquid metal.

As was mentioned above, Beeley⁽⁷⁰⁾ and others

have shown that an increase in the superheat results in an increased interface temperature.

In the present investigation, it was found that, in the case of vertical castings (Figs.62-71), increased superheat resulted in a greater development of the columnar growth and a coarsening of the central crystal zone. This may be explained as follows:-

The effect of increasing the superheat of the molten metal is to increase the thermal content of the casting as well as to eliminate some of the "potential nuclei". This, in turn, results both in the head taking longer to freeze over and the block longer to solidify. Thus the number of solid or "free" crystallites fed into the block during the early stages of solidification is reduced and the columnar crystals develop to a greater length from the mould wall before their growth is restricted. Furthermore, since the number of "potential nuclei" in the molten metal is reduced, the number of crystals formed in the interior of the casting will be reduced. This in turn results in the individual crystals developing to a greater size before being restricted by neighbouring crystals.

In some cases the "free" crystals are not sufficiently numerous to stop the growth of all columnar crystals, with the result that there is a gradual transition between the columnar and equiaxial zones. This is illustrated by casting SHA6 (Figs.66,71).

While the degree of superheat is low, numerous "potential nuclei" are present and the thermal content of the casting is smaller. The head of the casting freezes over rapidly, with the result that "free" crystallites are formed and tend to settle from the head, adding to those which have already developed from the "potential nuclei" in the block itself. Since this action probably occurs before solidification is far advanced into the block from the mould walls, the columnar growth is interrupted and restricted resulting in the formation of a nuclear or fine equiaxial structure. The lower the superheat the more abundant will be the "free" crystallites, the smaller will be the crystals and the greater will be the restriction to the columnar growth. This effect is illustrated in casting SHA2 (Fig. 63, 68) by the sharp demarcation between the columnar and nuclear zones of crystallization.

In the superheat series of castings (Figs. 64, 52, 65, 66), there is an interesting variation in the position of the equiaxial crystal zone in the central portion of the casting. In casting SHA3 (Fig. 64), poured at a low superheat, the equiaxial zone extends down the side of the block, in contact with the columnar crystals; in castings SHA5 and SHA6 (Figs. 65, 66), poured at higher superheats, the equiaxial zones are situated in the bottom portion of the casting. The reason for the coarsening of the structure has already been discussed but no satisfactory explanation of the change of position of the zones can be suggested.

Falling Crystallite Theory.

The theory of the gravity settling of crystallites during the solidification process has been adopted by a number of investigators to explain the crystal structure of steel castings. The earliest application of the theory was to explain the conical zone of purer nuclear/equiaxial crystals found in the lower regions of steel ingots^(8a) and castings⁽⁶⁰⁾. Further applications of the theory were made in papers by the Melting and Metallurgical Committee of B.I.S.R.A.⁽⁵⁷⁾ and by West^(16,58), to account for the existence of a zone of nuclear crystals in the bottom half of horizontal castings, and for certain variations in chemical composition within the castings.

Essentially the theory suggests that "free" crystallites, which form ahead of the main solidification fronts from the mould walls, can settle through the still liquid portions of a casting towards the bottom.* The discussion of differential solidification on page indicated a possible mechanism whereby such "free" crystallites can be formed ahead of an advancing columnar growth. They will be formed of a purer material than the liquid from which they grow, and will also be purer than the average composition of the columnar growth by virtue of the fact that the columnar growth takes place in liquid which becomes increasingly segregated as solidification proceeds.

* It may be shown, by the application of Stokes' Law, that a crystallite of 0.01 m.m. diameter would have a settling

velocity of 1"/minute, assuming the specific gravities of liquid and solid steel to be 7.0 and 7.5 respectively.

In the B.I.S.R.A. report⁽⁵⁷⁾ it was shown that the upper half of a horizontal casting is typically columnar, while the structure of the lower half is generally partly equiaxial and partly nuclear. The reason why the two halves differ in structure is that columnar crystals can grow downward from the top wall unrestricted by the presence of "free" crystallites, whereas columnar growth from the bottom surface tends to be inhibited by the formation of "free" crystallites "in situ" and by the settling of other crystallites from the upper portion of the casting. If the superheat is high, the number of crystallites formed is limited, (Page 175) and the lower half of the casting tends to be mainly equiaxial, whereas, with a low superheat, many more nuclei are present, so that large numbers of crystallites are formed and a predominantly nuclear type of structure is obtained.

Although columnar growth from a vertical face is not restricted by crystallites formed "in situ" it is, nevertheless, influenced by those that settle from the upper portions of the casting and by the relative positions of the other solidification fronts. For example, columnar growth from the portion of the vertical face near the bottom of the casting is liable to be restricted by the presence of a mass of "settled" crystallites; on the other hand, towards the top of the casting, columnar growth from the vertical walls is

relatively unrestricted, the "free" crystallites having settled towards the lower portions of the casting.

To illustrate the settling theory, typical examples of nuclear crystallizations and segregation obtained in the present investigation are considered below.

(a) A conical pile of nuclear crystals is found at the base of the vertical block El (Fig.15, Section V'-V"), extending from the bottom of the feeder (apart from a thin layer of chill crystals) vertically into the block. The chemical composition (Table V) shows that the nuclear crystal zone is considerably purer than the other crystal areas and the sulphur print (Fig.18) indicates that the areas of segregation are limited to the equiaxial zone of crystallization in the area of the casting last to freeze. This is in agreement with the view that the purer "free" crystals have settled towards the bottom of the casting, leaving the coarse segregated equiaxial area denuded of nuclei.

Composition of Nuclear Zone:	C	0.36 -	0.38 %
	S	0.039-	0.042%
	P	0.042-	0.044%
Composition of Block: (excluding Nuclear Zone)	C	0.39 -	0.46 %
	S	0.044-	0.049%
	P	0.047-	0.049%

(The greater segregating tendencies of S and P are also illustrated by these results).

A similar conical pile of nuclear crystals is observed in the base of castings WX1 and WX3 (Fig.56) where it is apparent that the mass of settled crystallites has

restricted the columnar growth from the bottom surface of the mould. Towards the top of the "pile", that is to say the part formed during the later stages of solidification, the influence of the settling effect is less and the columnar growth is longer.

(b) The results of chemical analysis of the different crystal zones in the horizontal casting C1 (Fig.8 and Table III), confirm those published in the First Report of the B.I.S.R.A. Melting and Metallurgical Committee⁽⁵⁷⁾ and later by West⁽⁵⁸⁾. ^{that} Fig.8 shows /there is a zone of columnar crystals extending from the upper face towards the centre of the casting, whilst the central position consists of a segregated equiaxial region which merges into the nuclear zone at the bottom. This latter transition suggests that as solidification proceeded, most of the crystallites developing from the nuclei in the central portion, and entering the central region from above, settled into the bottom of the block; relatively few nuclei or crystallites were left to develop in the central region, so they were able to grow, comparatively speaking unhindered, into equiaxial crystals. As would be expected, the average composition of the nuclear crystals is purer than that of the columnar and equiaxial : this is shown by the following results:-

	C	S	P
Nuclear Crystals	0.38%	0.044%	0.045%
Columnar Crystals	0.43%	0.045%	0.049%
Equiaxial Crystals	0.41%	0.045%	0.046%

Both the sulphur prints of the horizontal castings C1 and C2 (Figs.10 and 27 respectively) and to a certain extent the etched structures (Figs.9a and 26a respectively) show that the areas of segregate are restricted to the central equiaxial portion of the casting, i.e., the part of the casting last to freeze.

(The pipe present in casting C1 (Fig.9a) is of interest in that it illustrates the relatively loose network of equiaxial crystals obtained towards the end of solidification).

The transverse sections CE and CH (Fig.8) illustrate the effect of the settling action of the crystallites on the columnar growth from the vertical side walls. In the upper part of the casting, where growth is unhindered, the columnar growth proceeds from the wall to the centre of the casting, or until restricted by the columnar growth from the upper mould wall. On the other hand in the lower half the side wall columnar growth is considerably limited by the mass of "settled" crystallites. Similar structural effects can be observed in other castings, particularly in sections B2H (Fig.22), C2H (Fig.25), D2E, D2H (Fig.28).

(c) Analysis of the nuclear zone in the inclined block, casting E1 (Fig.5 and Table II), shows that the crystals that have settled towards the bottom during the earlier stages of solidification are purer than those formed later. This is in agreement with the "settling theory". The crystallites that

develop during the later stages of solidification, although possibly forming ahead of a solidification front, do so from a partially segregated liquid, particularly when the nuclear zone occupies the major portion of the casting. For example, considering casting B1 again, it will be found not only that the lower nuclear crystals in sections BE and BH are purer than those nearer the top of the nuclear pile, in the same sections, but also that those in the section nearer the bottom of the casting (BE) are purer than those towards the head (BH). This is because the block solidified directionally from the lower end towards the head. The analyses are tabulated below:-

		C	S	P
Section BE.	Upper Nuclear	0.39%	0.044%	0.046%
	Lower Nuclear	0.37%	0.039%	0.045%
Section BH.	Upper Nuclear	0.42%	0.044%	0.046%
	Lower Nuclear	0.40%	0.038%	0.045%

The sulphur print of casting B1 (Fig.7), and to a certain extent that of casting B2 (Fig.24), shows that the areas of segregate, which are assumed to be associated with the portions of the casting last to freeze, are concentrated in the central equiaxial regions in contact with the upper columnar growth.

In casting D1 (Fig.11 and Table IV) which is also inclined at 45° to the vertical the "settling effect" is complicated by the fact that the lower portion of the casting solidified after the upper, whereas the solid crystallites

continued to settle towards the lower portion at the bottom of the casting. As a result, the areas of segregate - illustrated by the sulphur print (Fig.14) and the chemical analyses (Table IV) - occur in the lower central portion of the block and continue into the feeder head. The macrostructure (Fig.11, Section H' - H'') however, reveals a nuclear zone in contact with the lower face, the formation of which is in accordance with the principle discussed above.

The lower half of casting D2 (Fig.28) contains a zone of "settled" nuclear type crystals. This has restricted columnar growth from both the bottom mould wall and the lower portion of the side walls.

The areas of segregate are concentrated in the central equiaxial area in contact with the upper columnar wall growth.

MICROSCOPICAL EXAMINATION.

Specimens taken from representative crystal zones of several castings were examined microscopically. A brief account of some of the structures observed is given below:-

Casting A1. This was a top-fed vertical block (Fig.2) which had a nuclear structure throughout. This microstructure (Figs. 98 and 99) consisted of a more or less continuous network of ferrite (which seemed to outline the boundaries of the original austenite grains) enclosing areas of pearlite, which contained "feathery" ferrite.

Casting E1. This was a bottom-fed vertical block (Fig.15) and in its macrostructure there was a thin shell of chill crystals enclosing a zone of columnar crystals, with equiaxial crystals in the central portion of the block, and some nuclear crystals at the bottom. Fig.100 is typical of the microstructure of the chill crystals : except close to the surface, which was partly decarburized, the structure consisted of massive, and somewhat rounded, ferrite in a pearlite matrix that contained some Widmanstätten ferrite. In the columnar and equiaxial zones (Figs.101 and 102), there were networks of ferrite which seemed to outline the branches of the original dendrites : these branches were represented by areas of pearlite containing "feathery" and Widmanstätten ferrite.

Casting E2. This was the bottom-fed vertical block (Fig.31) that had an anomalous structure, it was all

nuclear except for an "island" of dendritic columnar and equiaxial crystals. The microstructure of the nuclear zone consisted of relatively coarse ferrite in a pearlite matrix (Figs. 103 and 104). In the dendritic zone, the microstructure seemed to consist, at first sight, of ferrite occupying the positions of the branches of the original dendrites, with pearlite filling the spaces between the branches (Figs. 105 and 106). As, however, the non-metallic inclusions were concentrated in the ferrite and absent in the pearlite areas, it is evident that the regions now occupied by ferrite were last to solidify, i.e. that the pearlite areas represent the positions of the dendritic branches and the ferrite represents the material that had filled the spaces between the branches. This is the reverse of the structure that one would expect to find because, owing to selective freezing, the branches must originally have contained less carbon than the solid deposited later in the spaces between them. This reversal of the expected arrangements is also evident in most of the other microstructures, e.g. Figs. 99 - 103 : in all these the non-metallic inclusions are associated with the massive or network ferrite, which must therefore represent the solid formed from segregated liquid.

This reversal of the ferrite/pearlite structural relationships was observed and discussed by West⁽¹⁶⁾, who concluded that there were two causes : (i) diffusion of carbon, during cooling from the solidification temperature to the

upper critical point, from the regions where the concentration of phosphorus in the austenite was high, i.e. from the segregated regions which were originally rich in carbon as well as phosphorus, into the regions where the phosphorus content was low, viz. the branches of the original dendrites, and (ii) the fact that phosphorus raises the A_{r3} temperature, so that in passing through the critical range, ferrite is precipitated first in the regions where the phosphorus concentration is highest, viz. the segregated areas between the branches of the dendrites. The structures described above confirm Wests' observations.

The only other feature of the microstructures of the castings that seems to merit comment is the fact that the microstructure of the nuclear zone of the "anomalous" casting E2 (Fig.104) is itself somewhat anomalous. It resembles the typical structure of a chilled zone (Fig.100) much more than that of the nuclear zone illustrated in Fig.99. Both the macrostructure (ignoring the small isolated dendritic "island") and the microstructure of the anomalous casting are consistent with the assumption that the metal was cast cold, e.g. by teeming through a cold stopper, but it seems that this explanation must be ruled out because 23 cwt. of steel had been teemed from the ladle before casting E2 was poured.

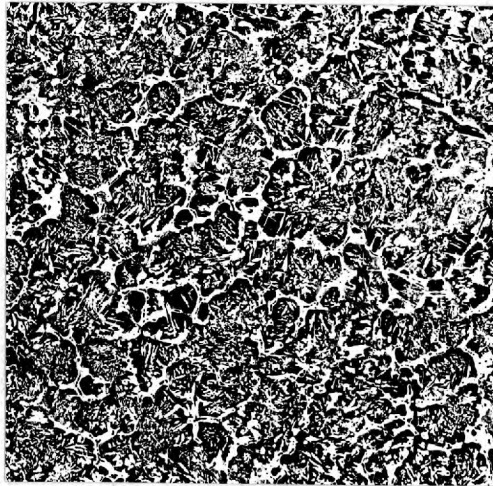


Fig. 98. Nuclear Crystals.
Casting Al.

x 12.

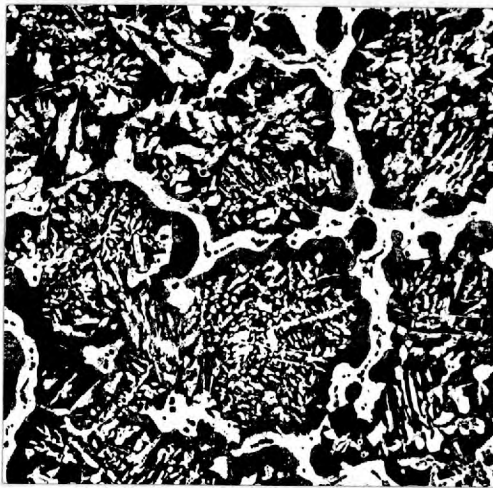


Fig. 99. Nuclear Crystals.
Casting Al.

x 35.

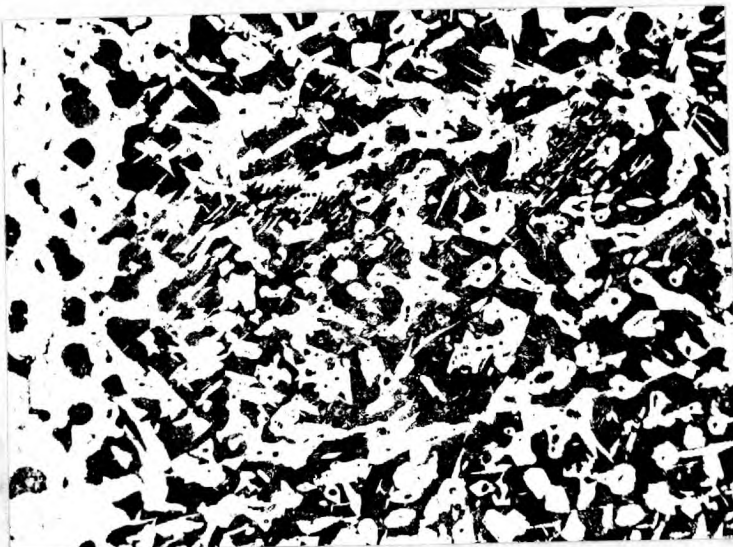


Fig.100. Chill Crystals
Casting El.
x 35.

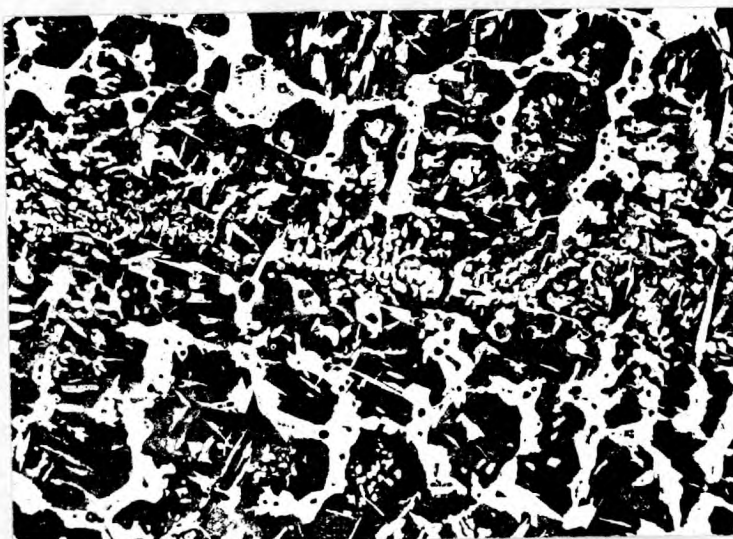


Fig.101. Columnar
Crystals
Casting El.
x 35.

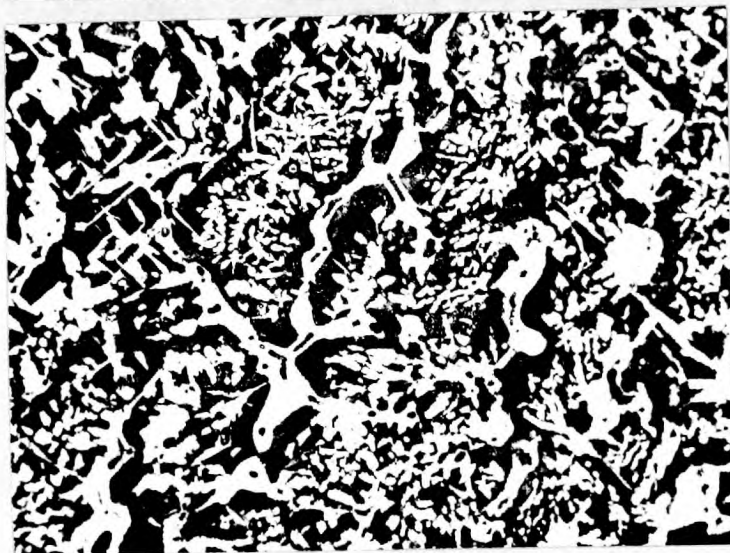


Fig.102. Equiaxial
Crystals
Casting El.
x 35.

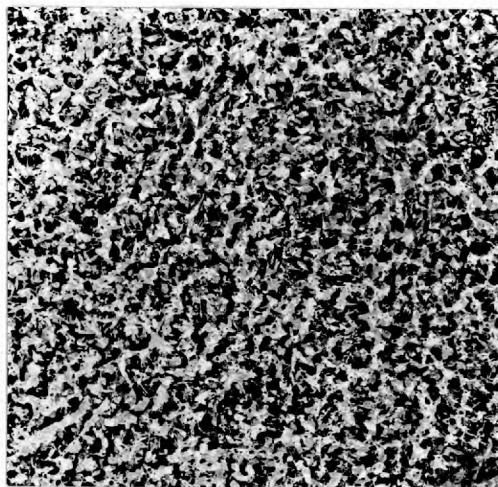


Fig. 103. Nuclear Crystals
Casting E2.

x 12



Fig.104. Nuclear Crystals
Casting E2.

x 35

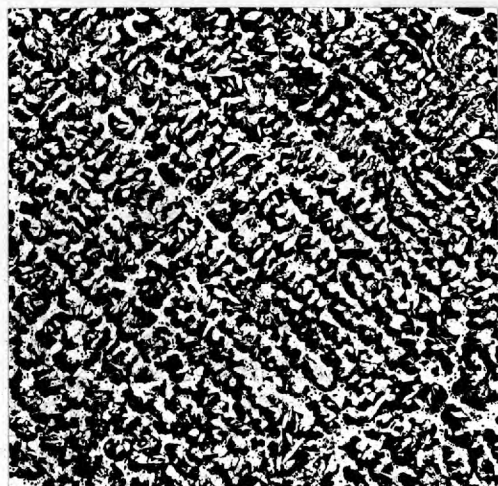


Fig.105. Dendritic Crystals
Casting E2

x 12

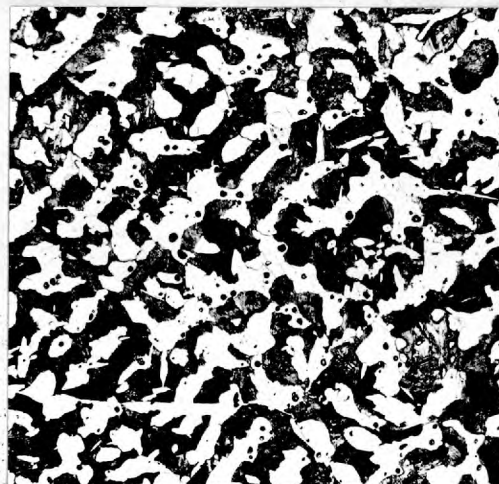


Fig.106. Dendritic Crystals
Casting E2

x 35

Summary of Conclusions.

(Note. The following conclusions refer to sand-cast steel blocks having roughly the following dimensions:- 5" x 3" x 2½" to 12" x 6" x 4".)

1. The mode of solidification of a steel casting is largely determined by the distribution of the centres of crystallization and the temperature conditions existing during the solidification process.
2. The distribution of the centres of crystallization and the temperature conditions are largely influenced by four factors:
 - (a) Superheat and casting temperature.
 - (b) Mould arrangement and convection currents.
 - (c) Size relationship between the casting and feeder head, and in particular the relative surface areas.
 - (d) The settling characteristics of crystallites under the influence of gravity.
3. The effects of superheating are:-
 - (a) To eliminate some of the potential nuclei in the melt, thus producing a coarser crystalline structure.
 - (b) To prolong the solidification time of a casting.
 The effect of the reduction in the number of "potential nuclei" and "free" crystallites results in a thicker columnar development from the mould walls.
4. The crystal structure of the block can be altered from a wholly nuclear form to a wholly dendritic type by varying the position of the feeder head. This effect is probably due partly to the influence of convection currents, but mainly to the settlement of crystallites.

5. The probable effect of the settlement of crystallites on the structure of the block may be summarized as follows:-

- (a) When a top-fed block is fed from a small head the head freezes over before solidification has far advanced in the block, with the result that free crystallites sinking from the head will tend to interfere with dendritic growth within the block and favour the formation of a nuclear structure.
- (b) When a top-fed block is fed from a large head, the head remains molten till a later stage in the solidification of the block with the result that free crystallites are not fed into the block: this will permit the equiaxial and columnar crystals in the block to develop till hindered by growth of neighbouring crystals formed "in situ". This will favour a coarse dendritic structure.
- (c) In a bottom-fed vertical block the only crystallites which can affect the structure are formed in the block and many of these tend to settle into the feeder at the bottom.
- (d) In intermediate castings between the top- and bottom-fed types the lowering of the feeder results in a transition from a condition favouring the feeding of crystallites from the head into the block to one favouring the settling of crystallites from the block into the feeder.

6. The independent crystallites which form during the early stages of solidification at a higher temperature are purer than those which form later in the process at a lower temperature. The former crystallites tend to settle towards the lower regions of the castings and frequently results in a zone of nuclear crystals purer than those crystals present in other areas which solidified from a more segregated liquid.

7. The influence of convection currents is probably strongest in bottom-fed and bottom-poured castings and weakest (or non-existent) in top-fed, top-poured types.
8. Typical structures of blocks fed in different ways may be summarized as follows:-
 - (a) Top-fed vertical block: A central area of fine equiaxial or nuclear crystals surrounded by a columnar growth from the mould walls, the relative extent of the different areas depending on the casting temperature and the dimensions of the block.
 - (b) Top-fed block inclined at 45° to the vertical: A columnar wall growth in the upper portion and a nuclear and reduced columnar growth in the lower regions. The central area shows a transitional equiaxial/nuclear form of crystallization.
 - (c) Side-fed horizontal block: In the upper half a columnar wall growth which extends to the geometric centre of the casting and a lower region of a nuclear crystallization. These two zones are separated by a transitional zone of nuclear/equiaxial crystals.
 - (d) Bottom-fed block, inclined at 45° to the vertical: A columnar wall growth from all the mould faces. The columnar crystals are more developed in the upper regions of the block. The central area between the columnar wall growths consists of equiaxial crystals and there are frequently areas of the nuclear type in contact with the columnar crystals along the bottom surface of the casting.
 - (e) Bottom-fed vertical block: An extensive columnar wall growth enclosing a central area of coarse equiaxial crystals, or extending to the centre in small castings.
9. The areas of segregate are associated with the portion of the casting last to freeze. The degree of

dispersion of the segregate is related to the crystal zone in which it is situated. If final solidification occurs by equiaxial crystallization throughout a large area of the casting the segregate areas are widely dispersed; whereas, if solidification is limited to a small area between two columnar growths the segregate areas are concentrated at the extremities of those crystals.

10. The ratio of the volume of surface area of a casting is an arbitrary measure of the solidification rate of a casting and the consequent crystal structure.
11. The addition of exothermic compounds to a feeder head has an effect similar to that of increasing the size of the head. Solidification is delayed and a coarser structure is obtained.
12. The temperature measurements show that during the early stages of solidification the superheat of the metal is rapidly extracted, the temperature of the metal throughout the casting cooling rapidly to an isothermal liquidus value. The metal in contact with the mould walls solidifies immediately after pouring, the temperature falling rapidly below the liquidus temperature.
13. Under certain circumstances it is possible for the sand/metal interface temperature to remain above the solidus for an appreciable period of time. This indicates that complete solidification is not always immediately obtained at the mould walls.

14. During the normal solidification of a steel casting there exists a band of freezing which corresponds to the interval between the liquidus and solidus isothermals. This band starts at the mould wall and progresses across the casting as heat is extracted by the mould walls. The width of the band of freezing is largely dependent on the rate of heat extraction and the temperature difference between the liquidus and solidus temperatures.
15. During the early stages of solidification there is a parabolic relationship between the distance solidification has progressed (as indicated by the end of the liquidus arrest) and time. However, the rate of solidification increases during the final stages and the parabolic relationship is no longer maintained. There is no parabolic relationship between the depth of complete solidification (as indicated by the solidus isothermal) and time.
16. The depth of solidification as indicated by "pour-out" tests corresponds to 5% solidification, in the case of columnar crystallization, within a casting, and 15% solidification in the case of equiaxial growth.
17. Columnar crystallization is accompanied by a narrow band of freezing and is favoured by rapid rates of unidirectional heat extraction under conditions of steep temperature gradients.
18. Equiaxial dendritic crystallization is accompanied by a wide band of freezing which frequently extends throughout

the interior of a casting during the latter stages of solidification. The equiaxial growth is favoured by slow rates of multidirectional heat extraction under conditions of shallow or negligible temperature gradients.

19. The formation of areas of nuclear crystals is favoured by:-

- (a) The settling characteristics of "free" crystallites, nucleated in the interior of a casting, which settle towards the lower regions.
- (b) The presence of "potential" nuclei within the melt. These are removed by superheating.

20. The columnar/equiaxial crystal transition found in some castings is probably caused by the restriction and interruption of the columnar growth by the presence of "free" crystallites ahead of the main solidification front, and is influenced to a large extent by the rate of cooling and solidification. When the rate of cooling is rapid it is probable that the columnar wall growth will extend a considerable distance from the mould wall before being interrupted by the "free" crystallites. (In small castings the columnar crystals may not be interrupted and may therefore extend to the middle of the casting). When the rate of cooling is slow the "free" crystallites will probably restrict the columnar development early in the solidification process.

The majority of the "free" crystallites are considered to have settled from a zone of solidification in the upper portion of the casting, such as the feeder head.

21. A comparison of the macro - and microstructure of representative crystal zones agrees with West's⁽¹⁶⁾ observations concerning the unexpected relationship between the primary structure and the distribution of ferrite and pearlite.

The non-metallic inclusions, which indicate the position of the solid formed from segregated liquid, are associated with massive in network ferrite, whilst the axes of the dendrites, which represent the solid formed during the early stages of solidification from a purer liquid, are associated with a pearlite and "feathery" ferrite. The reversal of the ferrite/pearlite structural relationship is probably due to two causes:

- (i) diffusion of carbon during cooling from the solidification temperature to the upper critical point, from the regions where the concentration of phosphorus in austenite is high into regions where it is low, and
- (ii) the fact that phosphorus raises the A_{r3} temperature, with the result that in passing through the critical range, ferrite is precipitated first in the regions where the phosphorus concentration is highest.

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APPENDIX I.Sulphur Printing.

The following procedure was found to give satisfactory sulphur prints of the large 12" x 6" x 4" sections:-

- (a) Steel surface carefully dry ground and cleared of grease with successive applications of benzene and, either ether or alcohol.
- (b) Light matte duplicating type photographic contact paper soaked for two minutes in 2% sulphuric acid.
- (c) Paper drained on an inclined glass plate until uniformly damp, but not visibly wet; excess acid removed with a small brush.
- (d) Paper carefully laid on the steel surface; rolling it on from one end so that minimum amount of air entrapped. By wiping the back of the paper with a flexible rubber squeegee intimate contact could be obtained between the steel surface and the photographic paper. (This action is facilitated if the width of the squeegee is greater than the width of the specimen). After laying the paper on the steel surface, the back of the paper should be moistened with 2% sulphuric acid as this is helpful in ensuring an even contact.
- (e) Paper carefully removed after a pre-determined time (about 30 - 40 seconds) and placed in a fixing bath of 20% hypo. After 15 minutes fixing, paper washed for one hour in water and finally air dried.

By employing the above method the first print taken was quite clear and satisfactory and it was unnecessary to take a second print, with accompanying loss of fine detail.

APPENDIX II.Etching Technique.(a) Oberhöffer's Reagent.

The etching technique finally employed after numerous tests was Oberhöffer's (Copper) Reagent; the composition of which is given below. This reagent was employed by West⁽¹⁶⁾ and was found to reveal both the fine and coarse structures present in a section of a casting without excessive pitting. After inking, the structures obtained were sufficiently clear to be photographed. In order to obtain a satisfactory etch on a 12" x 6" section the following procedure was adopted.

After sulphur printing the steel section was placed in a deep paraffin-wax lined tank, with the prepared surface uppermost, and immersed in fresh Oberhöffer's Reagent. The surface was covered to a depth of about half an inch. It was found necessary to agitate the solution and to swab the surface of the specimen to prevent uneven etching.

When the grinding marks had been removed, the section was taken from the solution, wetted and scrubbed lightly with a commercial cleaning powder. After drying with alcohol, the condition of the etch could be examined and if necessary the surface was re-etched until the required depth was achieved.

It was found that if the surface was only swabbed after immersion the macrostructure was rather obscure and unsuitable for photographing. Whereas if it was scrubbed it was found that the full detail could be seen, but lacked contrast necessary for photographic reproduction. This was overcome by applying a wash of Indian ink evenly over the scrubbed surface, and allowing it to dry. If it was then

carefully polished with emery paper, the crystal detail was clearly defined and of sufficient contrast to be photographed. (In removing the ink, smears were liable to form but these could be avoided by backing the polishing block with a yielding baize material).

Composition of Oberhoffer's Reagent.

Cupric Chloride	6.5	gms	Ethyl Alcohol	290	mls
Ferric Chloride	195	gms	Hydrochloric Acid (S.G. 1.16)	195	mls
Stannous Chloride	3.25	gms	Water	620	mls
					1 Litre

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