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THE EFFECT OF PRODUCTION VARIABLES ON THE PROPERTIES AND PERFORMANCE OF INGOT MOULD CAST IRON

by

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SYNOPSIS

THE EFFECT OF PRODUCTION VARIABLES ON THE PROPERTIES AND PERFORMANCE OF INGOT MOULD CAST IRON

It is extremely difficult to determine the effect of microstructure on the cracking resistance of cast iron by conventional mechanical tests. It has been established that a three point bend test may be used to rank irons of different microstructure in terms of cracking susceptibility.

Experimental test block material has been produced and the effect of various production practices on cracking resistance assessed using the bend test. Production ingot moulds have also been examined to determine the relationship between production process, microstructure, mechanical properties and ingot mould performance.

It has been established that high (>0.12%) phosphorus levels may cause premature mould cracking due to increased segregation of phosphide/carbide eutectic to the eutectic cell boundaries. This effect is enhanced by increased cooling rates and by increased residual levels.

Titanium acts as a graphitiser and offsets the deleterious effects of increased residual levels by increasing cracking resistance. Too high a titanium level (>0.06%) produces a weak iron making the mould prone to torn seats. The effects of titanium are altered depending on the redox conditions. A ductility trough has been found at 0.03 to 0.05% titanium.

Compacted graphite is not, necessarily, deleterious to performance. Its formation has been linked with high residual levels. Nitrogen has not been found to be contributory and has little effect on cracking resistance although levels above 0.014% promote pinholing.

High residual levels cause a marked decrease in cracking resistance by promoting a cell boundary network of carbides. Slow solidification rates promote this effect by increasing the amount of segregation.

Reduced casting temperatures cause a reduction in eutectic cell size with a corresponding increase in cracking resistance. In practice, reduced casting temperatures give reduced performance because of cold shuts in slab moulds and poor crazing resistance in small moulds.

PREFACE

This thesis is submitted in part fulfilment of the requirements for the degree of Doctor of Philosophy of the Council of National Acedemic Awards. The research described was carried out during the period September,1975 to September,1981 at the Sheffield Laboratories of British Steel Corporation and financed by them. No part of this thesis has been submitted for a degree at any other University or College.

The work described has been published, internally, for circulation within B.S.C. and two papers published externally. These are:-

"The Evaluation of the Cracking Resistance of Cast Iron by a Three Point Bend Test". The British Foundryman, 73, Part 3, March, 1980, pp61-66.

"Effects of Phosphorus and Foundry Stripping Practice on the Performance of Cast Iron Ingot Moulds". International Conference on "Solidification Technology in the Foundry and Casthouse", University of Warwick, 15th-17th September, 1980.

During the period of this work the author attended the following lectures, seminars and courses.

Sheffield City Polytechnic MSc in Metallurgical Engineering course on Solidification.

Institute of British Foundrymens' Annual Conferences, 1978 and 1979.

Symposium on "Electric Melting for Ferrous Foundries", 1976.

Seminar on "Cupola Melting Updated", 1977.

BCIRA Study Courses on "Nodular Iron", "Cupola Design, Operation and Control", and "Electric Melting".

International Conference on "Solidification Technology in the Foundry and Casthouse", 1980.

British Steel Corporation Internal Seminars on "Statistics", "Computers", and "The Future Production of Ingot Moulds and Bottom Plates".

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CONTENTS

The peak in tonnage of ingot mould consumption occurred in 1973,with over 500,000 tonnes being produced. Since that time,tonnage production has declined,though ingot moulds still represent a significant proportion of the total tonnage of iron castings manufactured in the United Kingdom,as the following table illustrates[1]:-

Year	1973	1974	1975	1976
Tonnage Ingot Moulds X1000	528	419	366	396
Total Tonnage Iron Castings X1000	3445	3190	3002	2963

1977	1978
311	298
2795	2689

The total ingot mould and bottom plate consumption within British Steel Corporation,given in their Annual Statistics for 1979/80 shows a similar trend,as follows:-

Year	1973	1974/5	1975/6	1976/7
Tonnage Ingot Moulds X1000	414.5	407.6	281.3	372.6

1977/8	1978/9	1979/80
244.7	266.9	204.5

This reduced demand for ingot moulds has arisen from three factors, principally,

i) a gradual improvement in performance over a number of years,

ii) a gradual reduction in steel production levels since the early 1970's,

iii) the introduction of continuous casting.

For the financial year 1980/81,the total liquid steel-make in B.S.C. was planned at 15m tonnes per annum(tpa). Even at this low level of production,12m tonnes were to be produced by the ingot route. This would require an ingot

mould consumption of 150,000-200,000 tonnes.

Not withstanding this reduced ingot mould consumption, the mould costs continue to represent a significant proportion of the conversion costs of liquid steel to the wrought product. A typical ingot mould costs around £220/t and,thus,the total cost to B.S.C.,for its ingot moulds,is around £40m p.a. Any improvement in the performance of ingot moulds would represent a considerable saving in production costs.

Under optimum conditions the interior surface of an ingot mould deteriorates to such an extent that the mould is discarded. However,in many cases moulds crack before this condition is achieved necessitating early scrapping. Ultimate failure by cracking has been accepted as the normal failure mode,particularly in large slab moulds. Complaints from the steelworks on premature failures due to cracking have accounted for annual claims of up to £500,000.

All ingot moulds used within B.S.C.,other than a small tonnage for experimental purposes,are manufactured within the Corporation's own foundries.Those currently producing ingot moulds are Fullwood in Motherwell,Dowlais in Merthyr Tydfill,Renishaw in Derbyshire and Stanton in Nottinghamshire. In addition,ingot moulds have also been produced at two of the Corporations steel foundries at River Don Works in Sheffield and Craigneuk,also in Motherwell. Other than Stanton,which is part of the Tubes Division of B.S.C. these foundries are part of the Forges,Foundries and Engineering Group,a profit centre currently separated from the Manufacturing Divisions of B.S.C. (A brief description of the method of mould manufacture used at each foundry is given

in Appendix 1,together with foundries that have recently ceased production.)

Typical ingot moulds produced and used within the Corporation and examined in this work are shown in Appendix 2.

The use of magnesium modified graphite cast iron for ingot moulds has increased over the last few years,but by far the greatest tonnage is still produced in conventional flake graphite iron. This investigation has,therefore,been concentrated on flake graphite iron,although reference is also made to modified graphite iron.

There is published evidence,(reviewed in the next chapter),to show the effects of foundry processing variables,such as chemical analysis,on mould performance. This is,however,often contradictory and there has been little or no attempt to determine the fundamental relationships between mould microstructure and mechanical properties, and the resultant ingot mould performance.

Therefore,the principal aim of this investigation was to study the effects of foundry production variables (e.g. chemical analysis,cooling rate) on the mould microstructure,properties and ultimately performance (particularly cracking resistance) of cast iron ingot moulds,with a view to optimising mould performance through correct control of foundry production conditions.

The method adopted was to develop a simple mechanical test which would provide a relationship between the microstructure and cracking resistance of cast iron produced as test blocks. It was intended that this would also establish a relationship between test block production conditions and

properties. Secondly,the information so derived was used in conjunction with actual ingot mould performance data of a number of different moulds produced under varying foundry conditions,to establish the required foundry conditions for optimum mould performance.

CHAPTER TWO

FACTORS AFFECTING INGOT MOULD PERFORMANCE

An ingot mould may be defined as a container into which liquid steel is poured, for the purposes of producing a solid ingot, which may, subsequently, be rolled or forged into the desired final product. Practically, the design of the ingot shape is based on solidification and yield considerations, so the mould is designed to contain the steel and to conduct heat away whilst it solidifies. During usage, within 5 to 15 minutes of teeming, the inside surface of the mould wall rapidly heats up to temperatures of around 700-900°C[2] followed by equalisation of the temperature gradients. The stresses developed within the mould wall, due to these temperature gradients, are represented by an equation of the form[3]:-

$$\sigma = \frac{\alpha E (t_w - t_2)}{1 - \mu}$$

where α is the coefficient of linear expansion of the mould,

$(t_w - t_2)$ is the temperature difference between the inner and outer surfaces

and μ is Poission's ratio.

Inspection of this equation gives some indication of the properties of an ideal mould material: The stresses generated during usage may be limited by utilising materials of low linear expansion, low Young's modulus and high thermal conductivity, so as to limit the thermal gradients within the wall.

The moulds, after use in the steelplant, are allowed to cool below 100°C and are reused. Mould reusage, therefore, is an important factor in the economics of the ingot route. During usage, the inside surface deteriorates through thermal fatigue, growth and oxidation. Under optimum conditions, the

processes continue to such an extent that,either the ingot cannot be stripped,or the resultant billet or slab surface quality is so poor that the increased dressing costs do not justify continued use of the mould. This failure mode is known as crazing. Frequently,however,moulds may fail before this condition is achieved,cracking being the most common cause of early mould failure.

2.1 MODES OF FAILURE

The surface deterioration mentioned above ultimately leads to rejection of the mould due to crazing. In addition to this ultimate rejection,other failure modes are commonly observed.

2.1.1 Cracking

Cracking occurs when the thermally-induced stresses are greater than the tensile strength or notch toughness of the material. It may occur within the first few lives of the mould or develop gradually during service. It is usually associated with rejection early in life and is important since,not only is there an increased cost to the steelworks, but the sudden cracking of a mould may cause a breakout of liquid steel which is hazardous to operatives.

Cracking manifests itself as vertical cracks(either on the broad or narrow faces of slab moulds),horizontal cracks, or corner cracks, and may be due to one or more of the following factors:

i) Poor mould design,

ii) Adverse usage conditions,

iii) Poor iron properties, ie. iron too brittle.

2.1.2. Torn Seats.

These defects are caused by welding of the ingot to

the mould wall and, on subsequent stripping of the ingot, part of the mould is removed with the ingot. This type of defect usually occurs where the iron is weak but may also be due to faulty teeming practice (see 2.1.6.).

2.1.3. Broken Lugs.

The breaking off of a lifting lug renders the mould unusable. Again, a common cause of this defect is weak iron, but it may also be due to mishandling.

2.1.4. Mechanical Damage/Stickers.

When an ingot is stuck in the mould, efforts to release it may result in permanent damage to the mould. Stickers may arise through use of a heavily crazed mould, distortion of the mould wall, or by the ingot becoming keyed in to part of the wall, for example, through a teeming defect or a torn seat. Moulds may also be damaged on removing the ingot from a cracked mould, where steel has penetrated between the crack faces.

2.1.5. Distortion.

The imposed thermal stresses may be higher than the yield limit of the iron so that plastic deformation may occur. Over repeated cycling, this may manifest itself as gross distortion of the mould. S.G. iron moulds(discussed below), are prone to this type of defect.

2.1.6. Teeming Defect.

If the steel stream impinges on the mould wall, the surface will be eroded and this may require the scrapping of the mould. This type of defect is most common in direct teem moulds, but may also occur in uphill teemed moulds, when ladles with running stoppers pass over the tops of the moulds.

2.1.7. Cold Shuts.

Cold shuts are horizontal line defects, usually apparent as horizontal cracks. They are caused by, mainly, low foundry casting temperatures or by slag entrapment in two-ladle casting systems.

2.1.8. Plucking

Plucking is the term used to describe where part of the face has been pulled off. This defect occurs, most often, in large forging moulds, where sheets of iron, up to 15mm. thick are pulled away with the ingot. Plucking usually occurs where there is a plane of non-metallic inclusions just behind the working face.

The method of failure of an ingot mould and the life achieved before this occurs,are determined by the steelplant usage conditions,ingot mould design and foundry production conditions. These factors will be considered in more detail.

2.2 STEELWORKS USAGE CONDITIONS

There are several important steelworks variables which may affect mould life[4];

i) 'Teem to strip' time; the longer this time,the more heat is abstracted from the ingot,and the hotter the mould becomes. Long 'teem to strip' times result in greater oxidation and growth,and crazing occurs at a greater rate,

ii) 'Strip to teem' time; if this time is short and the mould still hot when the next cast is teemed then temperatures in the mould are correspondingly higher,so that oxidation and growth processes will occur at a greater rate,so increasing the

rate of crazing. Similarly,if the mould is cold, particularly if it has been stored in icy conditions,then the thermal stresses generated,on pouring the ingot,will be greater,possibly resulting in cracking. There will be an optimum cooling period but this will vary from mould type to mould type,depending on section size. The forced cooling of moulds by water sprays may result in undesirable stress distributions so that cracking may occur.

iii) Teeming temperature; high steel temperatures result in higher thermal shock,steeper thermal gradients and higher mould temperatures,so that the likelihood of cracking and the incidence of crazing is increased.

Lewis and Memmot[5],from an analysis of casting pit statistics,showed that each occurrance of a ladle temperature above 1570°C reduced mould life by half a heat. They also showed that mould life was greatly affected by malpractices occurring early in its life. Using the method proposed by Lewis and Memmott for weighting malpractices thus occurring,the author carried out a statistical analysis of 25 tonne slab moulds and found that 50% of mould life varience was explained by consideration of the above three steelplant variables and only 10% by the chemical analysis of the mould iron. Thus,steelworks operating variables exert a great influence on mould life.

2.3 INGOT MOULD DESIGN

2.3.1 Mould Weight to Ingot Weight Ratio

If the mould wall is thin i.e. of light weight, then temperature and stress gradients, soon after pouring, are increased, so that the likely occurrence of premature cracking is increased. Similarly, the overall temperature of the mould will be higher, so that the processes leading to crazing will be accelerated. If the mould wall is thick, temperature and stress gradients are reduced and the bulk temperature is lessened, so increasing mould life, but, at the same time, increasing the iron consumption (measured as kg of mould iron used per tonne of steel cast). From this, an optimum mould:ingot weight ratio may be established, and has been found to be 1:1 [6]. This implies that, for example, a 6t ingot requires a 6t mould which can easily be produced, but for a 200t forging ingot, a mould of the required size is beyond the capacity of all foundries within B.S.C. In this case, a compromise is made, between optimum mould performance and actual mould production costs, so the mould weight is chosen to be of the order of 150 tonnes.

2.3.2 Ingot Mould Shape

It has been stated above that poor design can lead to premature cracking of moulds. A mould should be designed so as to abstract heat uniformly from the ingot surface. This will improve the solidification pattern within the ingot and lessen temperature gradients within the mould wall, hence improving performance. Numerous patents exist (see for example reference 7) showing how this may be achieved. However, the material properties of the mould are an important consideration, so any mould design must also

take into account this important parameter. For example, many mould designs call for the use of reduced corner thicknesses in relation to wall thickness[8]. This may result in increased stresses at the corners, so that optimum mould design is only achieved by the use of mould materials of sufficient strength.

2.4 FOUNDRY PRODUCTION PARAMETERS

Beside the obvious effects of analysis on performance, (which will be described in detail in section 2.6), the producing foundries may affect mould performance by causing variations in dimensions, integrity, surface finish and in mould properties by altering microstructure. The production of ingot moulds within the British Steel Corporation is regulated by a Code of Practice[9] based on many years experience, designed to define and control those practices which affect performance. A brief description of B.S.C. foundries and their methods of ingot mould production are given in Appendix 1. The Code of Practice is reproduced in full in Appendix 3. Those factors affecting ingot mould performance are discussed below.

2.4.1. Sand Practice

Traditionally, ingot moulds have been produced using dry sand practices and this system is still in use in a number of foundries. Typical sand systems employed use up to 90% return sand, being made up with 10% new, clay-bonded sand, plus additives of strengthening agents such as bentonite, cereals and water. After moulding, the sand moulds are stoved at temperatures of around 350°C for 12-15 hours.

More recently, several foundries have changed to the use of resin bonded sand systems for ingot mould production.

The systems are usually based on urea formaldehyde/furfuryl alcohol resins,catalysed by para-toluene sulphonic acid.

The Code of Practice[9] states that the aim of the sand practice employed is to produce a rigid,permeable sand mould,with adequate high temperature properties,which will, in turn,enable the production of a sound,dimensionally accurate ingot mould,free from surface defects.

Poor sand compaction may result in erosion,causing entrapped,non-metallic inclusions,and shrinkage defects through mould dilation (this is discussed in section 2.4.6). Poor compaction and inefficient painting may also result in burn-on and metal penetration,and if this layer is not removed,it would act as a thermal barrier during steelworks usage,so impairing mould performance.

The particular sand system employed affects the cooling rate of the casting,for a given sand thickness,for example,by variations in thermal conductivity and the effects of casting into a warm dry sand mould. This will effect mould performance and is discussed in section 2.4.7.

2.4.2. Dimensional Accuracy

The ingot moulds produced in the foundry should be dimensionally accurate. Misplaced cores may result in non-uniform wall thicknesses, which may increase the incidence of cracking. The dimensional tolerances obtained in ingot moulds have been presented by Tupholme and Wilson.[10].

Mould Wt.Range.	Wall Thickness	Ingot Cavity	Ingot Wt.
10t	±3%	±0.7%	±1.5%
10-20t	±5.5%	±2%	±1.5%
20t	±5.5%	±2%	±1.5%

Deviations outside this range may cause problems in

the steelworks with under- and over-sized ingots. Reduced taper may cause hanger cracking. In addition, swells on the internal faces may cause the ingot to be keyed in, giving difficult stripping, which may, in turn, cause mechanical damage to the mould. Dimensional accuracy is controlled chiefly by sand practice. If the sand mould is efficiently rammed, the tolerances defined by the pattern equipment should be obtained.

2.4.3. Casting Temperature.

Cast irons are poured in the temperature range, usually, 1200-1450°C, depending on section size and chemical analysis. For ingot moulds which are of eutectic composition and of heavy(>75mm.) section, casting temperatures usually fall in the range 1200-1300° C.

If the casting temperature is low, around 1200°C, there is an increased likelihood of the formation of shuts and laps, whilst at temperatures above 1300°C, there is a marked increase in sand burn-on, which may increase fettling costs. High casting temperatures also result in increased liquid shrinkage and increased danger of breakouts from the moulding tackle.

Metallurgically, Banks[11] has claimed that low casting temperatures of about 1200°C impair the properties of mould iron and reduce ingot mould life. The mechanism of this effect was not indicated, however. In practice, a casting temperature is, therefore, specified for ingot moulds, of 1250-1280°C.

2.4.4. Pouring Rates.

Whilst the mould is being filled, radiation from the molten metal may cause breakdown of the sand bond, with

subsequent washing away of the sand. Pouring rates should be, therefore, as fast as possible. The Code of Practice recommends 3t/min. and 6t/min. for moulds of less than and greater than 5t, respectively, and, generally, moulds are filled in two to four minutes.

2.4.5. Running Systems.

The Code of Practice states that running systems should be sufficient to allow the pouring rates given above and should be positioned to avoid impingement of the metal stream on to the sand.

Although fettling is not covered in the Code of Practice, it is well established that care should be taken when removing the runners by burning, the white iron structure thus produced is brittle and may initiate cracking in the mould body, so it is usual to leave a 25-35 mm. long stub.

2.4.6. Feeding.

The Code of Practice states that the mould should be fed sufficiently to obtain a sound, dimensionally accurate product. This will usually require the use of 4% feed metal.

Moulds require to be sound, since any shrinkage cavities may increase the likelihood of cracking and may cause operational problems in bottletop moulds. Shrinkage cavities may also act as thermal barriers.

Unlike steel, cast iron undergoes a solidification expansion due to graphite precipitation. When iron is cast at 1250°C, this expansion nearly counteracts the liquid contraction so that, theoretically, no feed metal is required. In practice, however, sand mould dilation occurs, and generally about 4% feeder volume is required for contraction. The sand mould should be sufficiently rigid to minimise

dilation. The use of furane resins,when properly compacted, produces extremely rigid moulds,so that the requirement for feeding may be eliminated,but riser heads must be used to take up any expansion and prevent irregular tops resulting from the pressure generated during solidification.

2.4.7 Cooling and Stripping

The Code of Practice requires that moulds should not be stripped until the casting is below 700°C and that the minimum sand thickness should be 125mm.

Cooling rates,after casting,have been shown to have an influence on subsequent mould performance. If the mould is stripped at 1000°C then the carbide/phosphide eutectic phase is still liquid (m.p.=960°C) and fast cooling rates can promote its solidification in the form of a eutectic boundary film,which has been shown to be deleterious to mould performance[12]. Fast cooling,through the austenite transformation zone,results in reduced pearlite inter-lammelar spacing,which,again,has been shown to be deleterious[12].

In practice,slow cooling is achieved by maintaining sufficient sand thickness and by stripping the ingot mould from the tackle only after full transformation,that is below 700°C. A sand thickness of 125-150mm has been found to give acceptable performances in moulds up to 15t in the case of dry sand systems. There is little data available on the optimum cooling rates to produce a satisfactory product. Research,however,is being given impetus by the use of furane resin systems. In order to reduce moulding costs,it is necessary to reduce the sand thickness to the minimum possible,whilst maintaining a satisfactory product.

In the case of 25t slab moulds it is known that Australian practice uses only a 150mm sand thickness, compared with 300mm for similar moulds in B.S.C. It is apparent, therefore that considerable scope for improvement is possible.

2.5 CHEMICAL ANALYSIS

Mould analysis affects mould performance through its effect on microstructure and hence properties. The two important microstructural parameters are the matrix and the graphite morphology.

The performance of ingot moulds has been related to chemical analysis by numerous workers (see for example references 5 and 13-18). Immediately it emerges that there is no one ideal chemical analysis. For example, some workers show that a high Si:Mn ratio (i.e. 2% Si and 0.5% Mn), that is, partially ferritic, is beneficial, whilst others show this to be detrimental, so that the particular analysis will depend not only on the mould type, but also on the usage conditions. Several general guidelines may be used, however; where a mould usually fails by crazing, an improvement in life may be obtained by increasing the amount of pearlite stabilizers, such as Mn and the trace elements and reducing the graphitizers, such as Si, Ti, and Al. This has the effect of increasing the volume fraction of pearlite and increasing its stability, so that growth and oxidation is reduced and the crazing is slowed down. When a mould fails by cracking, the analysis is adjusted in the opposite direction, and the amounts of the graphitizing elements are increased and the pearlite stabilizers reduced. This results in an increased volume fraction of ferrite, which decreases the strength of the iron but has been

found, empirically, to increase the ductility of the iron and, hence, reduce cracking.

It may be seen, immediately, that the principal effects of variation in analysis will be through changes to the ingot mould microstructure. Particularly significant are the volume fraction of pearlite and the graphite morphology.

A convenient non-destructive method of determining the morphology of graphite in cast iron is the use of ultrasonic velocity measurements. The technique utilises two probes placed across the wall thickness of the casting as an emitter and a receiver. An ultrasonic pulse is passed between the two probes and the time interval measured automatically. The pulse velocity may then be calculated from the known wall thickness. In ingot mould practice, velocities of 3.2km/s represent an extremely coarse graphite structure and 4.0km/s usually representing the finest achievable in the heavy sections involved.

Graphite structures of compacted, quasi-flake and spheroidal forms (see section 2.6) give ultrasonic velocities of the order of 4.3km/s, 4.8km/s and 5.2km/s, respectively.

The foundry Code of Practice does not recommend any specific microstructure but gives analysis limits. It states that the five major elements should fall within the range

C	Si	S	P	Mn
3.7-4.0	1-2	0.1max	0.15max	0.5-1.1

and that the recommended residual levels are as follows:-

N	Sn	Cu	Ni	As	Cr
0.007max	0.02max	0.3max	0.2max	0.01max	0.05max

Mo	Pb
0.01max	0.0005max

It recognises that these levels may require adjustment as further work is carried out.

The rationale behind the limits to analysis is detailed below.

2.5.1 Carbon

Carbon in hematite ingot mould cast iron is present in the form of flake graphite and also carbide in pearlite or in the cell boundaries. The flake graphite networks confer on the cast iron its suitability as an ingot mould material, by producing high thermal conductivity and resistance to thermal distortion. In addition, the high carbon content aids founding by producing a low melting point.

Carbon and silicon levels are usually maintained so as to provide a near eutectic composition, the carbon equivalent (defined as $C+\frac{1}{3}Si$) being held at around 4.3%. Hypereutectic irons produce Kish graphite on solidification, which is deemed to be undesirable. Low carbon contents, however, produce a lower thermal conductivity, and it is generally considered that the higher strengths of these lower carbon irons result in a more crack-sensitive material.

Carbon contents are, therefore, maintained at the highest level possible, consistent with economic furnace operation and are usually in the range 3.7 to 4.0%.

2.5.2 Silicon

Silicon acts as a graphitiser in cast iron, increasing the carbon equivalent, and is usually added to promote a eutectic composition. Silicon stabilises ferrite, in preference to pearlite. At high silicon levels, however, undercooled graphite is promoted and this is used in the heat

resisting Silal irons. In addition silicon in solution in ferrite is known to cause embrittlement above approximately 2.2%. Additions of FeSi alloys act as an inoculant, their effectiveness being increased by small amounts of Al, Sr, Ca, etc.

In ingot moulds, silicon levels are usually in the range 1-2%, variations within this range being made to improve mould performance. Generally, 1.0-1.4% silicon gives optimum performance in small, square moulds failing by crazing , levels above 1.4% being used in large slab moulds failing by cracking. This arises from its graphitising and ferritising effects ; higher silicon levels in this range apparently producing a more crack resisting iron.

Ingot mould foundries may make a late addition of FeSi alloy to the transfer or casting ladle, in order to bring the analysis within the required range.

2.5.3 Sulphur

In flake graphite cast iron, high sulphur levels are known to stabilise carbides. At low levels, difficulty in inoculating the iron may occur.

Sulphur levels in flake graphite ingot moulds are usually maintained at around 0.05%, and there is no evidence that variations in sulphur level affect performance. These sulphur levels may be readily achieved in basic cupola practice. Where acid operation is used separate desulphurisation is carried out, using lime or calcium carbide. 0.05% S may be achieved in electric furnace practice by the careful selection of the melting stock.

2.5.4 Phosphorus

Phosphorus levels in ingot moulds are maintained at 0.15% maximum. However, the third report of the Ingot Mould Sub-Committee[19] provided evidence to show that phosphorus contents in excess of those levels normally associated with ingot moulds (0.06% approximately) may be beneficial for the performance of small moulds (4t) . The range considered was 0.06 to o.23% phosphorus and the beneficial effect was enhanced by relatively low silicon levels (1.0-1.4%) . It has been shown,[20] that phosphorus levels of up to 0.4% may be employed, provided that the moulds are slow cooled, being stripped only after full transformation has occured.

2.5.5 Manganese

Manganese fulfils two roles in cast iron. Firstly, it fixes sulphur as MnS. It is considered that to avoid the formation of deleterious sulphide phases the required manganese content is defined by

$$Mn = 0.3 + 1.7 \times (\%S) .$$

Secondly, manganese is a powerful pearlite stabiliser. Normal manganese levels in ingot moulds are held in the range 0.6 to 1.1%, most moulds having nearly fully pearlitic structures.

Ingot mould foundries, producing mainly slab moulds, use 0.6-0.8% manganese and those producing small square moulds, use 0.8-1.0%. This is because higher manganese levels increase the crazing resistance of the iron but also increase the cracking suseptibility. In slab moulds, where cracking is usually a problem, manganese levels are, therefore, lower than in moulds failing by crazing.

2.5.6 Titanium

The reported effects of titanium on cast iron appear confusing, since it is claimed, by separate investigations, both to coarsen the graphite or refine it, producing under-cooled, Type D graphite. Titanium is a powerful graphitiser, it being claimed that 0.1 to 0.15% Ti is equivalent to 0.7% silicon [26]. Titanium has also been recommended for the graphitization of a white iron[27], despite the fact that it forms a stable carbide, Ti(CN). Comstock [26] suggests that titanium's graphitizing effect is due to deoxidation, although no mechanism was proposed.

Norbury and Morgan [28] observed the formation of under-cooled graphite by the addition of 1-2% of ferro-carbon-30% titanium alloy. Carbon dioxide bubbling after the addition was found to enhance the effect, whereas bubbling hydrogen through the melt produced coarse graphite. It was postulated that under oxidising conditions, liquid titanate inclusions were formed, which were non-inoculating whereas additions of silicon, calcium silicide and aluminium form solid nucleants in the melt. Hydrogen was assumed to act by reducing iron oxide from the titanate inclusions, thus increasing their melting point.

Dawson, et al[29], in investigating the effects of titanium and gas treatments, found that bubbling CO_2 or argon through the melt reduced the hydrogen content. Therefore, they attributed the effects of gassing treatment to the hydrogen content of the melt, high hydrogen contents producing coarse graphite, and low hydrogen levels aiding the formation of under-cooled graphite. Very low sulphur contents were also found to produce under-cooled graphite

and, more recently, the under-cooling effects of titanium has been attributed to its removal of sulphur from solution[30]. Previous work, within B.S.C. has considered the effects of titanium in ferritic irons of ingot mould composition and heavy section[23]. It was found that the graphite structures were similar in the range 0.016 to 0.074%, but at 0.197%, a few areas of under-cooled graphite were observed. A minimum tensile strength was found at about 0.04% Ti and it was suggested that this was due to initial removal of nitrogen from solid solution, in the ferrite, followed by solid solution strengthening of the ferrite by titanium. A subsequent investigation also showed this strength minimum to correspond to a minimum in the fracture toughness of the iron[24].

Since hot blast cupola melting was introduced at Fullwood Foundry during the 1960's, considerable experience has been gained as to the effect of titanium in the range 0.01-0.12%, on the performance of large slab moulds. This experience showed that increasing titanium levels in cupola-melted iron was accompanied by coarsening of the graphite, with an associated reduction in the strength of the iron. Thus, levels of 0.1% Ti gave an extremely weak iron, which failed at low life by tearing. Alternatively, low level titanium (0.01%) irons were brittle and failed prematurely, by cracking.

Krasovitskii, et al[25], postulated that the beneficial effect of titanium on ingot moulds was due to removing nitrogen, oxygen and hydrogen from solution, and also by increasing the amount of ferrite in the microstructure.

2.5.7. Nitrogen.

The solubility of nitrogen is dependent on metal analysis, temperature and partial pressure of nitrogen above the melt, and the following equation has been derived[31]:

$$\log\%N = 1000/T - 0.86 - 0.24(\%C + Si/4) + 0.015\%Mn + \tfrac{1}{2}\log Pn.$$

Where P_n is partial pressure of nitrogen.

For irons of typical ingot mould composition, the solubility of nitrogen in liquid iron is, approximately, 0.003%-0.004%. Nitrogen levels at melt-out are usually considerably in excess of these values, levels as high as 0.0265, although unusual, having been reported[21].

One source of such non-equilibrium values is the use of steel scrap in the cupola charge materials. It has been demonstrated that as the proportion of steel scrap in the charge is increased, then the nitrogen content rises[32]. Holding of the iron at temperature will give a reduction in nitrogen level towards the equilibrium value[31].

Additional sources of nitrogen are from carburising materials, sand binders and mould paints[33].

The effects of nitrogen in cast iron have been reviewed by several authors(see, for example, references 32-37). Increased nitrogen contents are claimed to promote pearlite formation, which may lead to an increase in tensile strength, and may give, also, white iron structures.

In heavy sections, levels above 0.008% are claimed to promote the formation of compacted graphite. This graphite form is claimed to be undesirable in ingot moulds, as it lowers thermal shock resistance, so the Code of Practice recommends a maximum nitrogen level of 0.007%. In lighter engineering castings, high nitrogen contents promote

the formation of blowholes and fissures.

Experimentally, the effects of nitrogen have been investigated by adding nitrogen bearing materials to the melt[38]. Typical additions include sodium ferrocyanide, urea and ammonium salts but the most commonly used additive is calcium cyanamide. This is usually added with soda ash as a flux and has been used for the deliberate increase of nitrogen to produce compacted graphite in ingot moulds. These methods have been reasonably successful in increasing nitrogen levels, though a factor often overlooked is that their use may introduce other elements. For example, the use of $CaCN_2$ may be expected to increase the calcium content of the melt. Calcium itself is known to produce graphite modification, so that when changes in graphite morpholgy occur, by the addition of $CaCN_2$, calcium or nitrogen may be responsible.

2.5.8. Residual Elements.

Besides those elements detailed above, numerous other elements may be present in ingot mould iron in the form of impurities. The Code of Practice suggests that the following trace element target levels be adopted:

Sn	Cu	Ni	As	Cr	Mo	Pb
0.02 max.	0.3 max.	0.2 max.	0.01 max.	0.05 max.	0.01 max.	0.0005 max.

Besides these elements there are many others which are not normally analysed.

The presence of lead in flake graphite iron has been shown to cause the formation of "Widmanstatten" graphite, which weakens and embrittles the iron. The presence of hydrogen has been demonstrated to promote this effect. Usual lead contents in cupola-melted ingot mould iron are around 0.0002%.

Of the other elements specified in the Code of Practice, all are pearlite stabalizers, whilst Cr and Mo also stabilize carbides. The actual residual levels achieved in practice, in ingot mould iron depend on the raw materials used for melting and the following are typical values achieved in the mould foundries:

	Sn	Cu	Ni	As	Cr	Mo
Renishaw*	0.006	0.106	0.057	-	0.08	0.016
Distington+	0.004	0.03	0.03	-	0.03	0.015
Fullwood+	0.005	0.035	0.02	-	0.05	0.016
Dowlais	0.015	0.07	0.05	0.018	0.04	0.005
Stanton	0.010	0.12	0.05	0.019	0.04	-

* Duplex practice.

+ Cupola practice.

It may been seen that there is considerable variation between the foundries, trace elements being generally lower at Distington, as a greater proportion of pig iron in the charge is used.

The effects of residual elements have been studied in detail in small moulds by means of regression analyses at the author's laboratories. Most data was available for Renishaw moulds, where, because of the different manufacturing processes available (i.e. cupola, electric or duplex) (see appendix 1), the resultant wider analysis range produces more statistically significant results. The following regression equations have been generated:

Atlas moulds $Life=55.2+28.8(Si)-288.7(Ni)$

B120 moulds $Life=133.6-378.4(Cu)$

$Life=116.6+35.24(S)-1140(P)+760.5(Ni)$

$Life=1546-34.77(Si)-205.0(Ni)$

610 OT/WB moulds	Life=217-28.9(C)-14.1(Si)-18.7(S) -178.6(P)+409.6(V)+91.4(Cu)
WEU 100X moulds	Life=82.89-320.1(P)-2591(Sn)+299.1(Ti) +269.7(Ni)
	Life=63.10-74.35(P)+438.7(Mo)
F-Type mould	Life=129.6-440.2(P)-138.9(Cr)+426.4(V) +326.8(Ni)

Details of the mould types are given in Appendix 2.

The significant factor to emerge from these equations is that there is an apparantly beneficial effect of increased residual element levels for some moulds (e.g. 610 OT/WB) but a detrimental effect for others (e.g. Atlas). It is considered that where moulds fail by crazing, increased levels result in pearlite stabalization, so that life is improved but for moulds failing by cracking, reduced levels result in increased cracking resistance.

Data on the effect of residual levels on the performance of large slab moulds is very limited, but a similar effect on cracking resistance may be expected.

2.6 MODIFIED GRAPHITE INGOT MOULDS

With the introduction of spheroidal graphite cast iron in the early 1950's, attention was drawn to its suitability for ingot moulds. Its potential benefits for ingot moulds arise from its greater strength and ductility and, hence, improved cracking resistance and greater thermal shock, thermal fatigue, oxidation and growth resistances, which will lead to an improved crazing resistance. In practice, these considerations may lead to an increase in life of 1.5 to 2 times[39].

Because of the higher ductility and lower thermal conductivity of SG iron(25-40 $Wm^{-1}K^{-1}$ compared with 42-60 $Wm^{-1}K^{-1}$ in flake iron), initial trials with SG iron moulds showed that they were prone to distortion to such an extent that the ingot could not be stripped. It was found that this problem could be overcome by modifying the mould design, usually involving a reduction in corner wall thickness[8].

Partial treatment with magnesium or treatment with magnesium and titanium in combination, results in an intermediate type of graphite, termed by various workers as "compacted", "vermicular" or "quasi-flake (Q/F)" graphite. For purposes of clarity, this report will refer to 'compacted graphite' as that form of thickened flake found in heavy castings and attributed to nitrogen, 'vermicular graphite' as that form occurring in low sulphur, fast cooled irons, and to 'quasi-flake' graphite, as that form produced by magnesium treatment.

The mechanical and physical properties of quasi-flake iron are intermediate between flake and SG iron[40]. It offers, therefore, improved cracking and crazing resistance compared with flake iron, but its higher thermal conductivity (38-54 $Wm^{-1}K^{-1}$) means that the degree of distortion is less than in SG irons. In practice, this means that design modifications are not normally required.

The widespread use of SG and quasi-flake ingot moulds is still limited, due to their comparative newness and mould design problems. Therefore, the bulk of ingot mould production is of flake graphite iron.

2.7 EVALUATION OF MECHANICAL PROPERTIES.

Although mould cracking is affected by mould design

and steelworks usage conditions, and is more prevalent in large slab moulds, metallurgical factors are thought to play a major part. Thus, the controlled cooling recommended in the Foundry Code of Practice Document[9] is used to reduce the cracking tendency in service. The control[9] of carbide-stabilizing, residual elements is, similarly, expected to minimise the cracking tendency.

However, there is still some confusion concerning the effects of certain metallurgical parameters on cracking resistance. Premature cracking has variously been attributed to nitrogen, carbide/phosphide networks, compacted graphite and prior austenite dendrites. Even in the case of the established parameters (cooling rate, pearlite content), there is little quantative data available and the metallurgical examination of prematurely failed moulds still relies heavily on subjective interpretation.

It is, therefore, important to establish a relationship between the material properties, iron microstructure and mould performance. It should be realised, however, that ingot mould microstructures are heterogeneous, and graphite and matrix morphologies are likely to vary substantially, depending on the location within a given mould. Furthermore, a premature crack, once initiated in an embrittled part of the mould, may propagate through the tougher bulk of the iron. This is illustrated in the case of the early removal of top plates, during manufacture, to expose the top surface of the hot casting, which produces a locally embrittled iron, which, in turn, may lead to failure of the mould. So it would not be possible to relate the premature failure to the bulk microstructure of the mould or of a test block.

Nevertheless, mould microstructure does, qualitatively, affect mould performance, through its effect on the material cracking resistance. It is, therefore, necessary to relate the cracking resistance of iron to the microstructure on a laboratory scale initially, and then, in turn, relate the appropriate property to mould performance, either on ingot mould samples or on suitably produced test blocks.

Grey cast irons fracture in a tough or brittle manner, depending on the analysis and microstructure. However, the quantification of the ductility of cast iron is difficult by conventional mechanical tests, as in the tensile test, the elongation or reduction in area is low-of the order of 1%, even for a relatively ductile flake graphite iron. Similarly, the Charpy impact values are of the order of 5 J or below, on unnotched specimens.

It has been shown[41] that the toughness of cast iron may be characterised by a modified fracture test, using an unfatigued CKS-type specimen. However, these test pieces are expensive to machine, the tests require non-standard testing equipment and the results require skilled interpretation. A simpler and cheaper test has been carried out[42] using a three or four-point bend test. Using a 9.5mm square specimen at 5.7 mm centres, on a Houndsfield Tensometer, Banks[42] demonstrated the embrittling effect of stripping test blocks above the upper-critical temperature, a practice known, in some cases, to cause premature, ingot-mould failures.

Banks analysed the data in terms of the areas before and after maximum load, on a load-deflection plot, these areas reflecting, respectively, the energy to initiate fracture and the energy to propagate through the specimen.

This method of analysis has been used in the testing of steels, to determine the suseptibility to lamellar tearing, using a test piece similar to a modern CKS specimen[43]. This test has become known as the U.S. Navy Tear Test, and has also been used to assess austenitic S.G. irons[44]. Three point bend tests have been used, as well, more recently, in assessing the suseptibility of welded steel to lamellar tearing, using a 25.4 X 12.7 mm specimen at 101.6mm centres. The data was analysed[45] in three ways, the following parameters being measured: the deflection to maximum load, which was considered to reflect the ease of fracture initiation; the deflection from maximum load to a point corresponding to half maximum load, which reflects the ease of fracture propagation; and, finally, the total area under the load-deflection curve, which was considered to represent the work done in initiating and propagating the crack through the notched ligament.

The bend test technique has, therefore, been evaluated in order to define its suitability for comparing the cracking susceptibility of ingot mould material.

In order to quantify the effects of processing variables on cast iron microstructures, properties and ingot mould performances, two approaches have been used. Firstly, laboratory material has been studied, and, secondly, data on the performance of ingot moulds has been analysed. Material for the evaluation of a suitable mechanical test technique has been drawn from both sources.

3.1 EVALUATION OF THE BEND TEST.

On the basis of the published literature, the bend test was employed as a simple but a relatively sensitive technique for comparing the cracking susceptibility of ingot mould material. It was decided to produce a series of experimental irons, with a range of graphite and matrix microstructures, in order to determine if the bend test was sensitive to such relatively minute changes in microstructure. If successful, the test could be used throughout the investigation as a method of ranking the cracking susceptibility of both the experimental irons and also production ingot moulds.

A cast iron test attachment supplied by Industrial and Educational Services was used (fig. 1). The specimen employed was a 150 mm long X 12.5 mm dia. cylindrical machined bar, which was bent under three-point loading on 127 mm centres. The attachment was connected to an Instron Universal testing machine to produce a continuous load-deflection curve. The crosshead movement was set at 10 mm/min. with a chart speed of 200mm/min. Three bend test samples were tested for each treatment/analysis condition to establish that results were consistent and reproducible.

3.1.1. The Effect of Graphite Size in Flake Graphite Iron.

Bend test samples were prepared from a series of blocks of similar melt analysis but with differing titanium and gas treatments to produce a range of graphite sizes. Microstructures of the five samples are given in Fig. 2, and full details of their method of production are given in section 3.2.2.

3.1.2 Effect of Matrix Microstructure

Bend test samples were prepared from a fully pearlitic iron. The test blocks were then annealed, sub-critically, at 700°C for 48 and 96 hours, to produce 60 and 100% ferrite, respectively, and further bend test samples prepared.

3.1.3 Effect of Carbide Phases in Flake Graphite Irons

Bend test samples were prepared from two fully pearitic test blocks, one of normally low phosphorus and one with 0.3% phosphorus. The production of the test blocks is detailed in section 3.2.1. The high phosphorus test block contained substantial amounts of carbide/phosphide eutectic segregated to the cell boundaries.

3.1.4 Effect of Graphite Morphology

Bend test samples were prepared from quasi-flake and spheroidal graphite castings, both types being carbide-free and fully ferritic. A sample was also taken from a mould with a compacted graphite structure with approximately 50% ferrite.

3.1.5 Effect of Carbides in Quasi-flake Irons

Bend test samples were prepared from a quasi-flake mould which contained a substantial quantity of eutectic carbides. The sample was then annealed at 950°C for 12 hours and further bend test samples prepared. These

specimens were found to be carbide-free.

3.2 EXPERIMENTAL CASTINGS.

Unless otherwise stated, the experimental castings were produced by a standard method. This was as follows:

1. Melt out a charge of pig iron in a 250kg capacity, high frequency induction furnace.
2. Trim analysis(additions of titanium were made by inclusion with the cold charge).
3. Superheat to 1400°C.
4. Tap out 125kg as required, into ladle.
5. Pour moulds at 1250°C.
6. In the case of split melts, hold the remaining iron at 1400°C, trim analysis and then tap.

Two mould sizes were used, principally, to produce a 255 mm. dia. block, weighing 125kg and a 305 mm. dia. block, weighing 250kg. Moulding box sizes varied, but in all cases, the minimum sand thickness was 150 mm. The moulds were produced in furane sand and were closed topped. No feeder heads were employed.

3.2.1. Phoshorus.

Melts were produced, using a charge of 100% Bremanger pig iron, or 50% Bremanger and 50% Warner pig irons, to simulate different residual levels (designated 'high' and 'low' levels for comparative purposes). Typical analyses of these pig irons are given in Table 1. 125kg test blocks were cast at high (0.3%) and low (0.06%) phosphorus levels. Half the blocks were fast-cooled, from above 1000°C, by stripping before 1½ hours.

Tensile and bend test specimens were prepared from the centre of each block. Microstructures were determined

from the ends of the fractured bend test specimens.

3.2.2. Titanium

The experiments were divided into three parts.

(a) The Effect of Titanium and Gassing on Light Section Castings.

Approximately 200kg of Bremanger pig iron (Table 1) was melted out, with additions of 1.0% Mn, 0.09% Cr and 0.05% Ni to ensure a pearlitic matrix. 20kg lots were tapped into a small handshank at 1400°C and hydrogen or carbon dioxide bubbled through for two minutes, after an addition of nominally nil, 0.1 or 0.2% Ti. A 150 mm. dia. open topped test block was then cast. Three controls were cast at the same Ti addition rates, but without gas bubbling. Bend test specimens were prepared from each test block.

(b) The Effect of Titanium and Gassing on Heavy Section Castings.

250kg melts of either 50% Bremanger and 50% Warner pig irons, or 100% Bremanger pig iron, were superheated to 1400°C and tapped into a ladle. The melts were treated with carbon dioxide or hydrogen by bubbling for 5 minutes and the control melts were left untreated. Then 250 kg test blocks were cast at 1250°C. Specimens for bend tests and tensile tests were prepared from the centre of each block.

(c) The Effect of Scrap Condition.

Since the literature showed that gassing may effect iron properties, the previous experiments being designed to test this, it was considered that the physical condition of scrap in electric melting may have a significant effect; for example, if the scrap was covered in iron oxide or not.

Accordingly, the use of clean scrap and used ingot mould scrap was investigated.

Hematite scrap was obtained from Renishaw Foundry (typical analysis shown in Table15) in the form of runners, which were subsequently shot-blasted to remove any adhering sand and dirt, and broken, used ingot mould scrap. Two 250kg melts for each scrap type were made, with nominal additions of 0.03% and 0.05% Ti. An addition of 0.2% graphite was added to the dirty scrap test blocks, to compensate for in-service decarbonization and any carbon oxidation during melting. The melts were superheated to 1400°C, then the furnace tapped and 300 mm. dia. test blocks cast at 1250°C and allowed to cool to below 500°C, before stripping; bend test, tensile test and metallographic specimens were prepared from the centre of each block.

3.2.3. Nitrogen.

Experimental melts were produced in a 250kg induction furnace, based on a charge of 50% Bremanger pig iron and 50% Warner iron, and two 125kg test blocks were produced from each melt. Treatments included $CaCN_2$ with soda ash flux, and, as controls, similar additions of Calsiloy were made. (This is a proprietory alloy containing 58% Si and 14% Ca.) High nitrogen ferro-manganese was used as a furnace addition, since it would produce, solely, an increase in nitrogen. N_2 gas was also bubbled through a melt at 1600°C, for ten minutes, to determine whether nitrogen pickup from this source was possible, similar to cupola operation and porous plug treatments.

All other melts were superheated to 1400°C, and all test blocks were cast at 1250°C. One block, from the FeMn

and N_2 treated melts, was stripped above 1000°C, all others were allowed to cool below 700°C before removing the sand jacket.

After stripping, the blocks were sectioned across the diameters and examined for gas holes: from the centre of the blocks, bend-test and tensile specimens were prepared and tested in the usual way. Microspecimens were taken from the fractured ends of the bend test bars.

3.2.4. Residual Element Levels.

250 kg. melts were produced in an induction furnace, based on a charge of 50% Warner pig iron and 50% Bremanger pig iron (Table 1). Three residual levels were selected, low levels produced with no further additions, this being similar to the Distington analysis (see Section 2.5.8.), a high level with an analysis similar to the maximum in the Code of Practice, and an intermediate level. The medium and high levels were produced by the addition of ferroalloys to the furnace.

Four types of test blocks were utilised, as follows;

(i) 255 mm. diam.

(ii) 255 mm. diam. with a 25 mm. insulating sleeve surrounding the cast.

(iii) 305 mm. diam. with a 25 mm. insulating sleeve.

(iv) 355 mm. diam. with a 25 mm. insulating sleeve.

Cooling rates of the castings were measured at the metal/sand interface, using Pt/Pt-Rh thermocouples with 10 mm. diam. alumina sheaths, packed with alumina powder.

All the castings were cooled to below 600°C, before removal of the sand. Bend test and tensile test specimens were then machined from the centre of each block. Specimens for microstructural examination were prepared

from the fractured ends of the bend test pieces.

3.2.5. Casting Temperature.

Two melts were made, using a charge of 50% Bremanger pig iron and 50% Warner pig iron (Table 1), in the 250kg induction furnace, both melts being superheated to 1400°C. From the first melt two 125kg batches were tapped and cast at 1250°C and 1200°C; from the second melt a first batch of 125kg was tapped and cast at 1300°C and a second batch was tapped onto 0.2% Fe-Si and cast, after approximately five minutes, when the temperature had dropped to 1250°C.

Blocks, 230 mm. dia. by 330 mm. high, were cast and then stripped after 24 hours, when the surface temperature was below 600°C. Four bend test samples were prepared from the centre of each block. Microstructures were examined on the bend test samples after fracture, and eutectic cell size was measured after etching in Oberhoffer's reagent. Pearlite interlamellar spacings were measured, by point-counting the volume fraction of resolvable pearlite, using a microscope objective of known numerical aperture[46].

3.2.6. Metal Superheat.

Two 250kg melts were made, using a charge of 50% Bremanger pig iron and 50% Warner pig iron (Table1). The first melt was superheated to 1300°C and held for 15 minutes. 125kg was then tapped and the first test block cast at 1250°C. The remaining iron was then superheated to 1400°C, and held for 15 minutes, tapped and cast at 1250°C. The second melt was superheated to 1500°C and held for 15 minutes. 125kg was, again, tapped and a third test block cast at 1250°C. The remaining 125kg was superheated to 1600°C, held for 15 minutes, tapped and cast at 1250°C.

The test blocks were allowed to cool below 600°C before stripping. Bend test and tensile test specimens were prepared from the centre of the test blocks and tested in the usual manner. Microstructures were examined on the fractured ends of the bend test samples.

3.3. INGOT MOULD DATA.

3.3.1. The Performance Of Large (25t) Slab Moulds.

From mid 1974 to mid 1975, the cupola burden at Fullwood Foundry consisted of 70% pig iron, 20% steel scrap and 10% foundry returns. The performance of moulds at Ravenscraig Steelworks was satisfactory, although the steel production levels were low. In November 1975 the burden was changed to consist of 53% pig iron, 17% steel scrap, 20% foundry returns and 10% broken moulds. This burden was the same as that used in Distington Foundry, the change being made to standardise cupola burdens at the two foundries.

Performance of moulds from this burden, most notably the Type 48, was very poor, although other mould types manufactured from the same burden also deteriorated in performance. (This mould is shown in Appendix 2). Most production at Ravenscraig over this period, was confined to the Type 48 and the main feature of the failure was the occurrence of premature cracking.

Despite attempts to improve performance, by carrying out repairs (particularly, bolting steel plates across the cracks), by June, 1976, the situation was critical, and it was decided, as an initial measure, to replace broken moulds in the cupola burden with pig iron, although the metallurgical justification of this action was considered

dubious. The resulting burden thus consisted of 63% pig iron 17% steel scrap and 20% foundry returns.

By November, 1976, however, it appeared that the performance of moulds made from this burden would still be poor and so it was decided to make an addition of ferro-titanium to the cupola burden. The cupola burdens at Fullwood have been changed a number of times, previously. This experience has shown that the addition of titanium may alleviate problems of mould cracking. However, when the Ti level exceeds 0.1%, severe problems are experienced with torn seats, due to a resulting weakening of the iron. Therefore, the addition of ferro-titanium to the cupola was limited to give a final mould titanium content of 0.05-0.06%. Since that time, performance has been satisfactory.

The production and performance of the 48-Type mould, for use at Ravenscraig, has been investigated by, firstly, a detailed examination of production practices at Fullwood and Distington. This, however, revealed no differences which would explain the poor performance. Secondly, an examination directed at determining the metallurgical cause of the poor performance of moulds produced, using the broken mould burden, was carried out. This latter course involved the following stages;

(i) Statistical analysis of performance for each burden and comparison with other suppliers.

(ii) Comparison of failure mode.

(iii) Statistical comparison of production details for each burden.

(iv) Examination of microstructures by optical and electron microscopy.

(v) Determination of mechanical properties, using bend test samples.

3.3.2. The Production and Performance of Forging Moulds.

River Don Works produce large forgings for the heavy engineering industries. The forging ingots, which may weigh up to 250t, are cast under vacuum. River Don Works, having their own foundry, commenced production of ingot moulds for their own requirements, in 1976.

The first mould (a 71"/68", 71t octagonal mould) was cast in November, 1976, and up to date, a total of eleven large moulds have been cast in the River Don Foundry.(A typical large forging mould is shown in Appendix 2). A number of performance problems have been experienced with these moulds, either due to cracking or face plucking. Because of the low rate of usage of large octagonal moulds at River Don (between 5-10 usages per year), the amount of information available is limited. However, the performance and production history of each mould has been examined, since it was thought that such an examination would offer additional insight in to the effects of very heavy sections and,hence, slow cooling rates, different usage conditions and also the effects of electric arc melting of cast iron.

3.3.3. The Effect of Phosphorus on Renishaw Ingot Mould Performance.

During 1974, the quality of steel scrap available at Renishaw deteriorated, both in size and trace element levels. A supply of railway chairs (40kg each with 0.75% to 1% phosporus and low residual content) became available and it was decided to charge these at a rate of 10-15% to the cupola, as a replacement for steel scrap, the balance being

ingot mould scrap.

The maximum phosphorus level resulting from this charge should have been approximately 0.16%, which was marginly above the Code of Practice maximum[9], but within the Ingot Mould Sub-Committee range[19].

Following initial trials, during which no adverse effect on performance was observed, the high phosphorus burden was initiated in July, 1974.

Concurrent with the change in burden, the pattern of operation at Renishaw was altered, due to a change in product mix and shift pattern. Previous daily practice had been to cast all ingot moulds before all other castings, in order to allow the maximum time before the top plates were removed, at about mid-night. In May, 1974, the practice was altered to include the pouring of up to three 17t castings before the ingot moulds were cast, but the time at which the top plates were removed was not delayed. On the 7 October, 1974, a three-shift system was introduced and Sunday working halted. This meant that, whereas moulds cast on Friday had been allowed to stand until Sunday in their sand jackets, Friday's moulds were now completely stripped by noon on Saturday.

From July to December, 1974, all moulds produced at Renishaw were produced from the new cupola burden. F-type moulds were produced for Tinsley Park, 610 OT moulds for Templeborough, B120 moulds for Aldwarke, WEU 100 and NEU 100 moulds for Round Oak, as well as a number of smaller orders for other customers. (Sketches of the principal mould types are shown in Appendix 2)

During December, 1974, a number of premature failures

of the F mould occurred at Tinsley Park, due to major cracking initiating from the base, the failures being confined, mainly, to high phosphorus moulds. Concern was expressed, not only at the increased steel costs, but also at the potentially serious safety hazard associated with the catastrophic cracking of a mould during teeming.

It should be emphasised that similar performance problems were not experienced at Renishaw's other ingot mould customers, at this time. However, the situation at Tinsley Park was so serious that as an immediate action, the phosphorus in the burden was reduced to its previous level by eliminating the railway chair scrap. Subsequently, all the high phosphorus moulds at Tinsley Park were heat-treated at 920°C for 10 hours and returned into service.

For each mould type, mean analysis and life over an approximately two-year period were calculated and multiple regression analyses of life versus analysis carried out; mean phosphorus levels and performance were then calculated on a monthly basis. Mean lives at various phosphorus levels and t-tests were calculated, to determine significant differences, and samples for metallography were taken from failed, high phosphorus, F-type moulds.

In order to observe the stripping practice at Renishaw Foundry, surface temperature measurements were recorded, using a Land surface pyrometer. Measurements were taken on three succesive Thursday nights, to compare the practice of each shift, the foundry working a three-shift pattern.

3.3.4. Examination of Prematurely Failed, Round Oak, WEU 100 Moulds.

Samples from five prematurely failed WEU 100 moulds

(appendix 2), made at Renishaw Foundry, were examined. The moulds were the subject of a complaint, having failed at low life, due to base cracking. Bend test specimens were taken from each mould at the base, and, approximately, 150 mm. from it. After testing, microspecimens were prepared from the fractured ends.

3.3.5. Material From Ingot Moulds For Routine Assessment.

Ingot moulds are usually cast with D-shaped lugs on one or two faces, to enable a check-chemical analysis to be carried out, if required. It was considered that the use of such a lug, if of sufficient size, could also be used to obtain bend test samples. Ten Renishaw and twenty Distington 610 WB moulds were produced with a large lug, from which two or three bend test samples were prepared. The properties of the samples were measured and compared with the ultimate mould performance.

In addition, Renishaw use a horngate in the running system. Prior to machining of the end faces, these horngates are removed and, usually, discarded. Ten such horngates were kept back and bend test samples prepared from them, and compared with the mould performance.

Since, if the bend test were to be adopted for foundry use, the measurements must be simple and quick, only the maximum load and a/b parameters were measured. Since there were a relatively large number of Distington lug samples, a statistical evaluation of life, analysis and all bend test parameters was carried out, by means of regression analysis. In addition, a correlation matrix of all variables was produced.

3.3.6. Comparison of Characteristic Foundry Material.

It has long been considered that the performance of moulds from certain B.S.C. ingot mould foundries is superior to others. In order to determine the reasons for these differences, 125kg test blocks were produced at Distington, Landore, Dowlais, Fullwood and Stanton (via cupolas), and at Craigneuk (via an arc furnace) from their usual ingot mould production. All the test blocks were allowed to cool to below 700°C, before stripping. Bend test and tensile specimens were then taken from the centre of each block and tested in the usual manner. Metallographic samples were taken from the fractured ends of the bend test specimens.

3.3.7. Effect of Casting Temperature.

In order, statistically, to determine the effect of casting temperature on mould preformance and the incidence of mould cracking, the performance of 1617 610 WB moulds and 197 L26 moulds, (see Appendix 2) produced at Distington Foundry, have been examined.

For the 610 WB moulds, the data were divided, initially, into casting temperature intervals of 10°C and life and failure mode for each group calculated. Mean casting temperature, analysis and performance were then calculated for each failure mode. T-tests were then carried out to determine significant differences between the crazed moulds and the other failure modes. A multiple regression analysis was then carried out for all moulds.

Because of the more limited data available for the L26 moulds (chosen since a higher proportion fail by cracking), the data was divided, initially, into groups

at casting temperature intervals of 25°C. The mean life and percentage failing by each mode, were then calculated for these groups. Mean life, analysis and casting temperature were then calculated for each failure mode and for all the data. T-tests were then carried out to compare the failure mode groups. Regression analyses were carried out for all moulds and for the different failure modes and casting temperature ranges.

4.1. EVALUATION OF THE BEND TEST.

4.1.1. Analysis of Results.

Banks has shown[42] that the principal differences between the load-deflection curves for tough and brittle irons is that, in the case of the latter, once the maximum load has been reached, the load drops more rapidly, as shown, schematically, in Fig. 3. This occurs because, in a brittle iron, only a small amount of energy is required to propagate a crack, so most of this energy is stored in the specimen as elastic strain energy.

A number of methods have been employed in order to measure the decrease in toughness, summarised in Table 2. A typical load-deflection curve is shown in Fig. 4, which also shows the measured variables.

A direct measure of the toughness of the iron is the area under the curve after maximum load (Area B, Fig. 4), which is a measure of the energy to propagate the crack. However, when examining a series of irons, it is noted that if these irons are embrittled, then the maximum load may be increased. This has the effect of increasing the area under the curve, so that a direct comparison of two irons of different strengths by measurement of propagation energies may be misleading.

To overcome this, the areas have been normalised (An and Bn). The ratios of the initiation energy to the propagation energy have also been used (A/B).

The typical load-deflection curve shown in Fig. 4, demonstrates the limitation of these techniques, due to the geometry of the system, in that the load never falls to zero. For the parameters A, B, A_n and B_n, given above,

the curve has been arbitrarily cut short 5mm from the zero line. To counteract this, the derived parameters $A_{\frac{1}{2}}/B_{\frac{1}{2}}$ and a/b have been measured which, although of less fundamental significance, are less dependent on the geometry of the system. The parameter a/b is also, physically, the easiest one to measure, since it does not involve time-comsuming area measurements, either by counting squares or using a planimeter.

The parameter (A+B) is that used by previous workers[45].

4.1.2. Effect of Graphite Size.

Typical bend test curves are shown in Fig. 5, and measured parameters in Table 3.

Reducing the graphite size from samples 5 to 3 (Figs. 2 and 5) tended to decrease the propagation energy (Area B, Table 3). This was associated with a slight increase in maximum load and initiation energy (A). The formation of under-cooled graphite, Samples 1 and 2 (Fig. 2) resulted in a marked decrease in propagation energy, such that, when the maximum load was achieved the crack propagated rapidly through the majority of the specimen. This clearly represents a highly brittle condition reflected in high A/B and a/b ratios. The normalised initiation energy (A_n) showed no clear trend, whilst the normalised propagation energy (B_n) showed a marked increase with graphite coarseness. The total energy to fracture (A+B) showed no systematic variation.

The tests show, overall, that an increase in graphite size is accompanied by an increase in toughness. Graphite size may be increased in the foundry by, for example, increases in carbon, silicon, titanium and section size.

These increases are known to cause a reduction in cracking tendency in ingot moulds, although the factors mentioned also affect pearlite and carbide content.

4.1.3. Effect of Pearlite:Ferrite Ratio.

Typical bend test curves are shown in Fig. 6 and measured parameters in Table 4.

As may be seen from Fig. 6 and Table 4, an increase in the amount of ferrite caused a slight reduction in maximum load (reflecting the tensile strength) but also caused an increase in the energy required for propagation of fractures. This has resulted in a lowering of the A/B ratios, which reflect an increase in toughness.

This, again, correlates well with common experience, in that a reduction in the pearlite content results in increased fracture resistance. The effect, however, appears to be less important than variations in graphite size.

4.1.4. Effect of Carbides in Flake Graphite Irons.

Typical bend test curves for low and high phosphorus test blocks are shown in Fig. 7, and the measured values in Table 5.

It may be seen that the increased amount of carbide/phosphide eutectic has increased the maximum load and, hence, initiation energy, but the propagation energy has been dramatically reduced.

This agrees with common experience, that when carbides are present, the iron becomes more brittle.

4.1.5. Effect of Graphite Morphology.

Typical bend test curves are shown in Fig. 8, for a hematite iron, a compacted graphite iron and a quasi-flake iron, and the bend test parameters are given in Table 6.

A ferritic S.G. iron when tested, did not fracture but deformed, plastically. The load-deflection curve was similar to the quasi-flake iron, although with a much higher deflection, up to the point when the test had to be stopped.

Analysing the results in terms of the a/b or A/B ratios, indicates that the modified graphite irons were more brittle, which clearly contradicts the results of conventional tests (tensile) and mould usage experience. This is due to a large increase in initiation energy (A), accompanied by a relatively smaller increase in propagation energy (B). This is shown in the normalised propagation energy values (B/M_L), which are essentially similar to the haematite iron, whilst the normalised initiation energy (A/M_L) is 2-3 times greater. In this example, the (A+B) parameter probably gives the best indication of the toughness variations which agree with experience.

The test, therefore, may be suitable for comparing quasi-flake and compacted graphite irons.

4.1.6. Effect of Carbides in Quasi-flake Irons.

Typical bend test curves are shown in Fig. 9 and the measured parameters in Table 7. It may be seen that the maximum load and initiation energies have been slightly increased by the presence of carbides, but that the propagation energy is reduced. This has resulted in increased A/B and a/b ratios.

4.2. EXPERIMENTAL MATERIAL.

4.2.1. Effect of Phosphorus.

Chemical analyses and mechanical properties of the test blocks are given in Tables 8 and 9.

Microstructures of the test blocks are shown in Figs.

10-13, where it may be seen that, in the case of the low residual melts, there is a substantial proportion of ferrite (85% in the slow cooled, low phosphorus test block), the amount being reduced by the fast cooling practice. The high phosphorus test blocks were more pearlitic than the low phosphorus blocks, but this is assumed to be due to the higher Mn level (0.73%, as opposed to 0.46%). The graphite was also coarser, probably due to the higher carbon content of 3.9%, compared with 3.75%, in the low phophorus block. Of particular note was that the phosphide/carbide eutectic in the high phosphorus blocks was in the form of a fairly uniform dispersion, with only a limited tendency to segregate to the cell boundaries.

The high residual melts, Figs. 12 and 13, were substantially pearlitic, with only a trace of ferrite being evident in the slow cooled blocks. The phosphide/carbide eutectic in the high phosphorus blocks tended to be more segregated to the cell boundaries than observed in the low residual melts. The high-phosphorus, slow cooled block contained a substantial proportion of compacted graphite, whereas the remaining blocks showed only traces of this form of graphite.

The bend test results (Table9) show an increase in brittleness at high phosphorus contents, high stripping temperature and high residual levels.

4.2.2 Titanium

Chemical analyses of the melts for the first two series are given in Tables 10-12. Tensile and bend test results are given in Tables 13 and 14. Tensile and bend test results for the heavy section (250kg), test blocks

are shown, graphically, in Figs. 14-17, and the a/b parameter for the light section (25kg) test blocks in Fig. 18.

The heavy section irons exhibited a strength minimum in the range 0.04 to 0.12% Ti (Fig. 14)for the gassed and untreated test blocks. This minimum tensile strength may be seen to correspond to maximum ductility in the irons (Figs. 15 to 17). The location of maximum ductility varies, however, with the gassing treatment.

The carbon dioxide treated, heavy section, test blocks showed an increase in brittleness from the residual level of 0.023% Ti, up to 0.055% Ti. The untreated, heavy section, test blocks showed a similar increase in brittleness from the residual level up to 0.035% Ti.

No hydrogen-treated, test blocks contained titanium analyses in this range. Above 0.12%. both the carbon dioxide and untreated test blocks showed an increase in brittleness, whereas no such trend was noted for the hydrogen-treated block. This behavior was also noted in the small 20kg test blocks (Fig. 18), which showed a steady increase in brittleness for the CO_2 and untreated casts. The hydrogen-treated cast, however, showed a reduction in brittleness at 0.18% Ti, but then an increase, again, at the highest titanium level.

The effects of gassing treatment on the microstructures of the small (20kg) test blocks are shown in Fig. 19. It may be seen that the hydrogen-treated blocks contained appreciably coarser graphite than the carbon dioxide-treated blocks. The non-gassed blocks showed a marked variation in structure at the low and medium titanium levels, although the graphite coarseness in the high titanium block was

intermediate between the H_2 and CO_2 blocks. The medium titanium H_2 treated block was slightly coarser than the low titanium test block, but in all other cases, the graphite increased in fineness as the titanium levels increased in the nominal ranges 0.02 to 0.2 to 0.4%.

Microstructures for the large (250 kg) test blocks treated with hydrogen, untreated or treated with carbon dioxide, are shown in Figs. 20, 21 and 22, respectively. The hydrogen and untreated blocks all showed an increase in graphite size up to 0.1% Ti, although this trend was less apparent for the CO_2 treated blocks. The highest titanium blocks showed a considerable amount of undercooled graphite with less being seen in the hydrogen treated blocks. The untreated blocks were, generally, coarser than the gassed blocks, but this may have been duu to their higher carbon and silicon contents.

There was a slight trend to increase the amount of ferrite within each set of blocks, an increase in titanium of 0.02 up to 0.1%, increasing the amounts of ferrite from approximately 5% to 25%, 2% to 15% and nil to 15% for the CO_2, untreated and H_2 blocks, respectively.

Analyses of the blocks produced, using clean and dirty mould scrap, are given in Table 15. Mechanical test results are given in Table 16 and typical micrographs are shown in Fig. 23. The microstructures are similar to those of the previous heavy section melts and the mechanical properties of the clean and dirty melts are similar to ungassed and CO_2 treated melts at similar titanium levels.

4.2.3. Nitrogen.

Treatments and stripping practices are summarised in

Table 17, test block analyses are given in Table 18, and changes in nitrogen and calcium levels, before and after treatment, are given in Table 19.

Evidence of gas porosity was seen only in cast RG 725A (2% $CaCN_2$-0.018% N), shown in Fig. 24. It may be seen that the porosity consisted of isolated pores up to, approximately, 25 mm. from the wall.

Mechanical properties of the test blocks are given in Table 20. Typical microstructures are shown in Fig. 25. It may be seen that the graphite morphologies were similar, with Type A flake graphite and traces of compaction in the eutectic cell boundaries, other than in cast RG 666 B (½% Calsiloy-0.007% N), which contained more compacted graphite. Matrix microstructures were, essentially, pearlitic, other than cast RG 752 B (2% Calsiloy); the higher proportion of ferrite in this cast is considered to have arisen from the much higher Si levels, brought about by the Calsiloy treatment.

4.2.4. Residuals.

Analyses of the melts are given in Table 21. Typical cooling curves are shown in Fig. 26 for the solidification interval and in Fig. 26b for times down to 600°C. It may be seen that the measured solidification temperatures varied considerably, but this is thought to arise from inaccuracies of the thermocouple: this effect on the measured times to cool down to, say, 700°C will be small.

Results from the mechanical property tests are given in Table 22. For convenience, the bend test a/b parameters are plotted in Fig. 27.

It is apparent that increased residual levels cause

a marked increase in embrittlement. The relative effects of section size (and, hence, cooling rate) appears to reverse, depending on the residual levels. Where these are high, fast cooled blocks are more ductile, but at low residual levels, the fast cooled blocks are more brittle.

Microstructures of the three residual level melts are shown in Figs. 28 to 30. It may be seen that as the residual levels are increased, there is a reduction in the amount of ferrite and also an increase in the proportion of compacted graphite. Carbide contents were measured by point counting on four blocks and the results are shown in Fig. 27. It is apparent that there is a high degree of correlation between the a/b parameter and the volume % of carbide.

4.2.5. Casting Temperature.

Chemical analyses of the test blocks are shown in Table 23, and mechanical properties of the test blocks, in Table 24, together with the pearlite interlamellar spacing and eutectic cell size. The effect of eutectic cell size on the $A\frac{1}{2}/B\frac{1}{2}$ bend test parameter is shown in Fig. 31.

The microstructures were all similar, consisting of flake graphite with a trace of compacted graphite in the cell boundaries. Pearlite contents were 85-95%. The pearlite interlamellar spacing showed no systematic variation. The eutectic cell size was reduced by a lowering in casting temperature and by the addition of FeSi. It may be seen that this effect improved the ductility of the irons.

4.2.6. Metal Superheat.

Chemical analyses of the test blocks are given in Table 25, and the mechanical test results in Table 26.

It is apparent, immediately, that no clear trend between properties and superheat temperature exists. Microstructures of the blocks were examined amd found to be similar, consisting of flake graphite with some graphite compaction in the cell boundaries and approximately 90-95% pearlite.

4.3. EXAMINATION OF MOULD PERFORMANCES.

4.3.1. Failure of Large Slab Moulds.

4.3.1.1. Performance Details.

Performances for the 48 Type moulds at Ravenscraig are given in Table 27, for each cupola burden at Fullwood. T-tests were carried out to determine which performance differences were significant, and these are given in Table 28.

It may be seen that the broken mould burden (53% pig iron, 17% steel scrap, 20% returns and 10% mould scrap-53/17/20/10) gave a significant reduction in life, compared with the original 70/20/10 burden (70% pig iron, 20% steel scrap, 10% returns). Performances were restored when broken moulds were replaced by pig iron (63/17/20 burden), this being maintained with the FeTi addition.

The distribution of lives about the mean for the 48-Type moulds are shown in Fig. 32.

The precise mould performances are complicated by the amount of mould repairs carried out. The number of repairs made to 48-Type moulds from the four burdens under consideration are given in Table 29. It may be seen that the incidence of repairs decreased from 51%, for the 70/20/10 burden to 20-24% for subsequent burdens. The higher level of repairs in the 70/20/10 burden might be expected to give an increased life over subsequent burdens, but the

precise effect is difficult to quantify. Since the level of repairing in the broken mould (53/17/20/10) burden was similar to the 63/17/20 and 63/17/20+FeTi burdens, the life increase may be attributed to improved iron properties rather than increased repairing.

It was noted, in Section 3.3.1., that, by November, 1976, it appeared that the 63/17/20 burden was not performing satisfactorily, but, however, final lives clearly do not substantiate this notion. This may be explained by consideration of performance and discard data and this is shown in Figs. 33-36. It may be seen that, initially, only low life moulds fail, the good life moulds failing only about six to seven months after the initial failures, and ten months after the burden has been changed.

4.3.1.2. Mode of Failure.

Using the Ravenscraig computer print-out, modes of failure for each burden were determined, and these are summarised in Table 30. For all burdens, it is apparent that most failures occurred due to vertical cracking. The position of these cracks is tabulated in Table 31, and it may be seen that the majority occurred on the broad side, except for the 53/17/20/10 burden, in which the majority were on the narrow side. As may be seen from Table 30 the incidence of cold shut defects for the broken mould burden (17.3%) was particularly high.

4.3.1.3 Analysis of Production Details

Fullwood production details are recorded on the Ravenscraig mould print out and these are summarised in Table 32. t-tests for significance were carried out between each burden at Fullwood and the results are given in Table 33.

For convenience the significant increases and decreases are summarised in Table 34.

Regression analyses to 90% confidence limits were carried out for life against production variables for each burden and for all Fullwood moulds. The results were as follows :-

70/20/10 Burden

Life = -84.21 + 38.38 x (%C) σ = 21.58
Life = 89.86 - 496.7 x (%S) σ = 21.32
Life = 56.51 + 274.9 x (%Ti) σ = 21.08
Life = -65.23 + 29.79 x (CEV) σ = 21.63

53/17/20/10 Burden

Life = -269.7 + 81.60 x (%C) σ = 20.77
Life = 75.17 - 546.2 x (%S) σ = 20.49
Life = 193.1 - 38.87 x (USV) σ = 20.26
Life = -231.1 + 63.73 x (CEV) σ = 20.89

63/17/20 Burden

Life = -14.57 + 52.31 x (%Si) σ = 21.06
Life = 96.45 - 706.5 x (%S) σ = 20.34
Life = -4.423 + 89.84 x (%Mn) σ = 21.30
Life = 40.29 + 568.0 x (%Ti) σ = 20.57
Life = 269.4 - 60.06 x (USV) σ = 11.55

63/17/20+Ti Burden

No significant relationships

All Fullwood Moulds

Life = -196.7 + 65.60 x (%C) σ = 22.49
Life = 35.04 + 17.62 x (%Si) σ = 23.06
Life = 88.60 - 593.1 x (%S) σ = 22.20
Life = 34.55 + 35.77 x (%Mn) σ = 23.09
Life = 48.12 + 341.1 x (%Ti) σ = 22.31
Life = 169.5 - 30.24 x (USV) σ = 22.61
Life = -185.0 + 55.57 x (CEV) σ = 22.50

(The equations only apply within the limits of analysis studied.)

Sulphur appears in the regression equations but its effect on life is questionable in the range considered (0.013-0.08%). It is considered that its effect is related to carbon in that high cupola basicities give high carbon

and low sulphur levels. Since increasing carbon improves life in the regression equations it may be surmised that sulphur will appear to give the opposite effect. A regression analysis of carbon against sulphur was carried out and it was found that :-

$$\text{Carbon} = 4.068 - 2.898 \times (\%S) \quad \sigma = 0.07723$$

this equation being significant to 0.1%. Since there is a high degree of correlation between carbon and sulphur, the effect of sulphur on life may be ignored.

4.3.1.4 Chemical Analysis

Major Elements

These have been considered in section 4.3.1.3.

Nitrogen

Nitrogen analyses for the 70/20/10, 53/17/20/10 and 63/17/20 burdens are given in Table 35. Although the general levels may be considered high there appears to be little or no change in overall levels between each burden.

Trace Elements

Trace element analyses for the 70/20/10 and 53/17/20/10 burdens are given in Table 36. Comparison of analyses for these two burdens on lug samples showed no differences.

4.3.1.5 Structural Examination

Samples from failed moulds were prepared for metallographic examination and etched in picral/nital. Typical microstructures are given in Figures 37-46. Included are mould numbers 48/585 (70/20/10 burden; 120 lives) and 48/758 (Distington manufacture; 65 lives). Figure 45 is from mould 48/023 (63/17/20+Ti burden), the sample being trepanned prior to entering service. Figure 46 is from mould 48/059 (63/17/20+Ti burden) which failed after two

lives due to a torn seat.

The microstructures were extremely variable and no clear trend was observed. The graphite form varied from coarse flake to highly compacted graphite. It is interesting to note that the Distington mould 48/758 (Fig. 38) which gave a satifactory performance contained compacted graphite Matrix microstructures were varied, but tended to be pearlitic. The matrices of the good performance moulds after failure (Figs. 37 and 38) were ferritic, but it may be assumed that, prior to entering service, they contained substantially more pearlite. The titanium-treated moulds were pearlitic (Figs. 45 and 46).

Mould No. 774 (Fig. 39) contained very fine graphite in a matrix of unresolved pearlite, with some phosphide/carbide eutectic. This is considered to have occured due to a high $\frac{Cr}{}$ content (0.22%, Table 36). No other examples of chromium contents at this high level, were observed.

The titanium-treated moulds showed a slight coarsening of the graphite flakes. This was reflected in a decrease in ultrasonic velocity.

4.3.1.6. Electron Metallographic Examination.

Five samples were chosen to represent high and low life moulds, with both compacted and flake graphite structures and a titanium-treated mould. The specimens were prepared by etching with 2% nital, carbon coating and extracting electrolytically, in 10% nital. Electron diffraction patterns were obtained, to establish the phase identities.

All the samples showed a bimodal size distribution of

TiN particles, of 0.15 μm and 0.03 μm. The relative abundance of each precipitate size showed no systematic variation with mould life although the compacted graphite samples tended to contain less of the fine (0.03 μm) TiN particles. The samples containing pearlite showed large particles of cementite after extraction.

4.3.1.7 Bend Test Properties

In order to investigate the iron properties of Fullwood moulds, samples were trepanned from mould walls after failure. Moulds were selected to give a range of performances and a titanium treated mould was included. This mould, number 48/059, had failed after two lives due to a torn seat but it was unclear whether this was due to poor iron properties or due to poor teeming practice.

The bend test results are given in Table 37, and actual curves for flake and compacted graphite irons are shown in Figs. 47 and 48, respectively. Microstructures of the samples have been given, above.

Examination of the bend test curves for flake irons shows that there was some difference between high and low life moulds, and it may be seen that the highest life mould failing by cracking, had the lowest a/b and A/B ratios (i.e. more ductile material) and vice versa. This effect, although showing a trend, may be enhanced by ferritisation during service, this having been shown to lower a/b and A/B ratios (Section 4.1.3.). The properties of the titanium mould, which failed due to a torn seat, fell in the range of the other three flake irons, so that it may be considered that the iron was not excessively weak, sufficient to cause this type of defect.

Less clear trends are apparent for the compacted graphite irons (Fig. 48), although the highest life mould possesses the lowest bend test ratios.

4.3.2. Production and Performance of Forging Moulds.

All the River Don Foundry ingot moulds were produced in the 90t arc furnace (Appendix 1), from a charge of cast iron borings (ex-Distington) and ingot mould scrap. The iron was cast into resin-bonded sand moulds. The analysis and production details for each mould are summarised in Table 38, and the performance details in Table 39. Full details of the mould manufacture and performance are presented elsewhere[47], however, the followong points may be highlighted:

(i) If the temperature of the ladle, on arrival in the foundry, is too high, then purging is carried out. This involves the injection of the iron with argon through a porous plug to give a thorough mixing. This might be expected to mix in the slag cover under some circumstances. Not all ladles required purging.

(ii) After stripping and fettling several moulds were found to contain non-metallic inclusions,particularly in the top. In all cases, except mould 71/68 No 1, this would be the wide end of the mould. Analysis of these inclusions showed them to consist of grains of nearly pure alumina in a matrix of an alumino-silicate. "Sandy" inclusions were found in the top face of several moulds during machining.

(iii) Several moulds ran to slag during pouring.

(iv) In a large number of cases when plucking occurred in the moulds it was associated with a malfunction during ingot teeming such as a full bore runner or a head breakout.

(v) When mould 91/86 No1 plucked after 9 lives the plucked surface was found to contain a light coloured phase. This was found, by scanning electron microscope, to be dendritic in nature and composed almost entirely of silica.

(vi) Microstructures of typical moulds are shown in Figure 49. It can be seen that they consist of extremely coarse graphite flakes in a predominantly pearlitic matrix. There is little evidence of carbides other than as noted in item (viii).

(vii) Bend test measurements were carried out on material from the heads of several moulds. The results were consistent with very ductile iron, typical A/B values being 0.6-0.9.

(viii) The plucked surface of mould 79/75 No1 was examined and at the surface was found to contain white iron. Further back from the surface there was a trasition to undercooled graphite and then to the normal coarse flake graphite.

4.3.3 Effect of Phosphorus on Renishaw Ingot Mould Performance

4.3.3.1 Renishaw Mould Performances

F Type Moulds at Tinsley Park

Average mould analyses for all moulds produced from January 1973 to January 1975 are given in Table 40. The monthly performances and phosphorus levels are shown in Table 41 and these are given graphically in Figure 50. It may be seen from Table 41 that, although the aim phosphorus content from August to December 1974 was 0.16%, in fact considerable scatter occured with minimum and maximum levels being 0.035 and 0.264% respectively.

Lives of individual moulds from this period are shown

in Figure 51. It may be seen that the lives of moulds with phosphorus levels of greater than 0.12% are markedly inferior. The mean lives in various phosphorus ranges are shown in Table 42 for all moulds produced during the above two year period.

Multiple regression analysis for all moulds was carried out and the following equation obtained :-

$$\text{Life} = 129.6 - 440.2 \times (\%P) - 138.9 \times (\%Cr) + 426.4 \times (\%V) + 326.8 \times (\%Ni)$$

95% confidence limits, $2\sigma = \pm 56$

this equation being valid within the limits of analysis given in Table 40.

Moulds cast between August and December 1974 were further split down, between those cast on Friday and those cast on Monday to Thursday. The mean lives at $\leq 0.12\%$ P and $> 0.12\%$ P are given in Table 43. t-tests were carried out but no significant difference between moulds cast on Friday and the remainder was found.

<u>610 OT Mould at Templeborough</u>

Chemical analyses of 610 OT moulds cast during January 1973 to January 1975 are given in Table 44 and monthly performance and phosphorus levels given in Table 45. These are plotted in Figure 52. Mean lives at various levels are given in Table 46 and mean phosphorus contents at given lives in Table 47.

Regression analysis of the moulds generated the following equation:-

$$\text{Life} = 116.4 - 18.7 \times (\%Si) - 89.3 \times (\%P) + 445.8 \times (\%V) + 458.6 \times (Ti) - 96.6 \times (\%Cu)$$

95% confidence limits, $2\sigma = \pm 45$

this equation being valid between the analysis limits given in Table 44.

B120 Moulds at Aldwarke

Mean chemical analyses for the period April 1972 to December 1974 are given in Table 48. Monthly performances and phosphorus levels are presented in Table 49 and are shown graphically in Figure 53. Mould performances with respect to phosphorus content are given in Table 50, and Table 51 gives the mean phosphorus content with respect to life.

Regression analysis gave the following equation :-

$$\text{Life} = 178.4 - 30.4 \times (\%Mn) - 37.0 \times (\%Si)$$

95% confidence limits, $2\sigma = \pm 44$

the equation being valid between the limits of analysis given in Table 48.

WEU 100X Moulds at Round Oak

Mean chemical analyses for moulds scrapped over the period September 1974 to March 1976 are presented in Table 52. Monthly performances and phosphorus levels with respect to date cast are given in Table 53 and Figure 54.

Multiple regression analysis generated the following equation :-

$$\text{Life} = 63.10 - 74.35 \times (\%P) + 438.7 \times (\%Mo)$$

95% confidence limits, $2\sigma = \pm 37$

this equation only being valid between the P and Mo limits given in Table 52.

Mean lives for a range of phosphorus contents are given in Table 54 and the average phosphorus content in specific life ranges in Table 55.

NEU 100 Moulds at Round Oak

Average mould analyses for moulds scrapped from September 1974 to March 1976 are given in Table 56. The monthly performances and phosphorus levels according to month cast are given in Table 57 and shown graphically in Figure 55. The mean lives in various phosphorus ranges are shown in Table 58 and average phosphorus contents at given lives in Table 59.

Multiple regression analysis calculated the following equation :-

$$\text{Life} = 81.98 + 205.6 \times (\%Cr)$$

95%confidence limits, $2\sigma = \pm 48$

this equation being valid in the range 0.01 to 0.18% Cr.

4.3.3.2 Metallographic examination

A typical fracture surface from a high phosphorus, prematurely-failed, F-type mould, is shown in Fig. 56. The structure exhibits a characteristic 'wire net' structure, which is known to cause embrittlement of the iron.

Typical micrographs are shown in Fig.57. It may be seen that there is a large interdendritic network of carbide/phosphorus eutectic which, in most cases, contained microcracks. The graphite form, although exhibiting a compacted structure, is not untypical of Renishaw Foundry ingot moulds.

4.3.3.3 Temperature Measurements.

Measured Temperatures on the top surfaces (as-cast) of moulds, immediately after removal of the top plate, for the three shifts, are given in Tables 60, 61 and 62. Temperatures of all the moulds were below 780°C, except for the F-type (two measured at 960°C) and the ECSC F-type (measured at 890°C, 880°C and 910°C). This latter mould is, essentially,

identical to the F-type, being only slightly modified externally, for use at River Don Works.

4.3.4. Examination of Prematurely Failed, Round Oak, WEU 100 Moulds.

Analyses of the moulds and their performances, are given in Table 63. The position of the bend test specimens are shown in Fig. 58, and the results given in Table 64. Typical bend test curves for mould 404 at the base and 150 mm. up, are shown in Fig. 59. Typical microstructures are given in Fig. 60-64.

It may be seen that the bend test parameters indicate that the moulds are substantially brittle compared with results from the preceding sections. The microstructures all show large amounts of carbide eutectic, and this is segregated to the cell boundaries, frequently being tendrillic in nature and forming a semi-continuous chain.

4.3.5. Material From Ingot Moulds for Routine Assessment.

Mould performances and bend test results for Renishaw horngates, Renishaw lugs and Distington lugs are given in Tables 65. 66 and 67, respectively. Microstructures of all samples were, essentially, similar, Typical examples are given in Fig. 65.

Although a/b parameters on several Renishaw horngate samples were relatively high, no mould failed by cracking. Two Renishaw lug samples gave high a/b values and one of these moulds cracked. Only one of the Distington moulds failed by cracking, and the a/b value for this mould was not high, compared with the others.

The chemical analyses and bend test parameters of the Distington moulds are given in Table 68. Multiple

regression analyses of life versus all variables (except casting temperature, where the data were incomplete) generated the following equation:

$$Life = 143.5 - 89.90X(\%Si) + 2043X(\%P) + 5.31(A_{\frac{1}{2}}/B_{\frac{1}{2}}) - 26.54(A/B).$$

The 95% confidence limits were $\pm$ 15 and the equation explained 88.5% of the overall mould life variance.

This equation appears to be contradictory, in that there are both negative and positive coefficients for the A/B and $A_{\frac{1}{2}}/B_{\frac{1}{2}}$ parameters.

The correlation matrix is given in Table 69. It may be seen that there are good correlations between the bend test parameters, maximum load, a/b, $A_{\frac{1}{2}}/B_{\frac{1}{2}}$, A and A/B, but none of these correlate, significantly, with life. It should be noted, however, that the correlation coefficients with life are positive. Parameter B correlates, negatively, with life. There are negative correlations of the bend parameters (except B) with C, Si and Ti, some of which are significant at up to the 99.9% level.

4.3.6. Comparison of Characteristic Foundry Material

Chemical analyses of the test blocks and mechanical properties are given in Tables 70 and 71. Typical microstructures for each foundry are given in Figure 66.

It may be seen that significant differences between the foundries exists with, in general, Distington giving the lowest a/b and A/B values and Craigneuk the highest.

4.3.7. Effect of Casting Temperature

For the 610 WB moulds mean lives for the various casting temperature ranges are given in Table 72 and these

are plotted in Figure 67. It is apparent that life increases with increasing casting temperature up to around 1340°C, after which data becomes limited. The effect of casting temperature on failure mode is given in Table 73 and the percentage failing by crazing plotted in Figure 68. A slight trend to a reduced incidence of crazing with increased casting temperature may be observed. For each failure group means of the production variables and life are given in Table 74 and t-tests against moulds failing by crazing in Table 75. A significant difference in casting temperature was found only for those failing with cracked corners whose casting temperature was 1275°C compared with that for crazed moulds of 1268°C. The lives of these two groups were 90.04 and 87.32 respectively.

Multiple regression analysis generated the following equation :-

Life = 0.137 x (Cast Temp°C) - 13.27 x (%C) - 25.2 x (%Si)

95% confidence limits, 2σ = 37.2

% variance explained = 9.3%.

This confirms the graphical observation noted above.

Performance and failure mode for each temperature interval for the L26 moulds are given in Table 76. No clear trend is apparent. Analysis, performance and casting temperature means for each failure mode are given in Table 78. Mean casting temperatures for each group are not significantly different.

Regression results were as follows :-

All Moulds - Life = 344.9 - 60.97 x (%C) - 46.93 x (%Mn)

2σ = 50 , Variance explained = 9.48% ,

Crazed Moulds- Life = 320.1 - 53.34 x (%C) - 56.00 x (%Mn)

$2\sigma = 51$, Variance explained = 8.78% ,

Cracked Moulds- No significant equations ,

Sticker Moulds- No significant equations ,

Moulds Cast 1250-1274°C - No significant equations

Moulds Cast 1275-1299°C - Life = 429.9 - 73.63 x (%C)
- 50.34 x (%Si)

$2\sigma = 49$ Variance explained = 14.46% ,

Moulds Cast 1300-1324°C - Life = 243.2 - 237.7 x (%Mn)

$2\sigma = 54$ Variance explained = 36.01% .

Clearly there seems to be no significant effect of casting temperature for this mould type with the relatively smaller amount of data.

5.1 EVALUATION OF THE BEND TEST

Ideally, the bend test should be able to rank materials in terms of cracking susceptibility, by means of a suitable and easily measured parameter which, in turn, should be possible to relate to the material microstructure and, ultimately, the mould performance, if failure occurs by cracking.

The main derived parameters for the different irons are summarised in Table 79. It may be seen, somewhat surprisingly, that, apart from the quasi-flake irons(Groups 5 and 6) and compacted irons (Group 10), the total energy for fracture (A+B) is relatively insensitive to material ductility, all samples exhibiting values in the range 1.2-2.7. Only in the case of the ductile quasi-flake iron were values as high as 10 observed. The parameter is, therefore, unsuitable for ranking irons of similar stucture, but may be used to highlight marked changes in toughness from one group of irons to another.

The remaining parameters (A/B, $A_{\frac{1}{2}}/B_{\frac{1}{2}}$, a/b) all reflect the relative sensitivity of the material to cracking susceptibility, high values being associated with brittle material. Although A/B can provide a useful ranking, its measurement presents problems because of the difficulty of obtaining an accurate measurement of the energy to propagate fracture. Although the ratios $A_{\frac{1}{2}}/B_{\frac{1}{2}}$ and a/b have no physical significance, they are much easier to measure and both increase as material cracking susceptibility increases.

The only exceptions to these observations were in the magnesium modified graphite irons, where increasing the degree of compacted graphite, with an associated toughness

improvement, also caused an increase in A/B ratios, indicating a more brittle material. This has already been discussed (4.1.5) and suggests that the A/B parameters are not suitable for ranking irons of widely different graphite forms, but the parameter may be used for comparisons of ductility of irons of the same graphite form.

Overall, the bend test appears to offer a simple and cheap technique, both for quality control purposes and as a research tool for predicting cracking tendency.

5.2 PHOSPHORUS

Examination of Table 9 shows that in the case of the high residual experimental melts, an increase in phosphorus content in the slow cooled blocks produced a marked increase in brittleness, the a/b ratio increasing from 3.5 to 14.3. A similar increase was also found to occur in the fast cooled blocks. The fast cooled blocks were also embrittled compared with the slow cooled blocks, a/b ratios increasing from 3.5 to 7.1 and 14.3 to 17.5 for the low and high phosphorus blocks respectively. The most brittle condition was represented by the high phophorus fast cooled test blocks.

In the low residual melts the fast cooled high phosphorus test blocks again represented the most brittle condition, but it may be seen that the increase in cooling rate had a greater effect on properties than changes in phosphorus levels. In fact the matrix was more pearlitic in the case of the high phosphorus test blocks and this has been shown to increase brittleness (see section 4.1.3), indicating that the real effect of phosphorus in this series was only slight.

The effect of phosphorus on the high and low residual melts was substantially different. The most notable microstructural variation, other than ferrite content, was that the phosphide/carbide eutectic in the low residual melts was dispersed, whereas in the high residual melts it was more segregated to the eutectic cell boundaries. It is apparent from this that high phosphorus levels will be most detrimental to cracking resistance when the phosphide/carbide eutectic is in the form of a cell boundary network. The test blocks have shown that this form is promoted by high trace element levels. This is considered to arise from the increased Cr and Mo levels in the cell boundaries due to segregation promoting formation of the carbidic phase.

The bend test results illustrate the importance of controlled cooling of moulds in the foundry. a/b ratios for low phosphorus casts were found to increase from 1.1 to 2.3 and 3.5 to 7.1 with fast cooling for low and high residual melts respectively. Comparing the phosphide/carbide eutectic form in the test blocks and in prematurely failed moulds (Figures 10-13 and 57), it may be seen that in the latter it is more segregated to the cell boundaries and also greater in volume. Comparison of Renishaw F-type mould analyses (Table 40) with those from the experimental test blocks (Table 8) shows that the Renishaw moulds contained substantially higher levels of the pearlite and carbide stabilising elements, Cu, Cr, and Mo than the high residual test blocks. This would indicate that the Renishaw moulds will be more prone to the deleterious effects of phosphorus and fast cooling than the test blocks, and also

explains the increased volume fraction of phosphide/carbide eutectic. The cooling rate on solidification will also play an important role since the slower the solidification rate, as in ingot moulds compared with test blocks, the greater the tendency for segregation at the head of the advancing solidification front and hence the greater the tendency to form a cell boundary network of phosphide/carbide eutectic.

It is clearly apparent from Figures 50 and 51 that the premature failures at Tinsley Park of the F-type moulds produced at Renishaw were due to phosphorus contents in excess of 0.12%. Table 42 demonstrates that the average mould performance dropped from approximately 120 lives with low phosphorus (<0.07%) down to 68 lives for high phosphorus (>0.131%) moulds.

The 610 OT, B120 and WEU 100 moulds showed drops in lives of 87 to 77, 107 to 103 and 65 to60 respectively (Tables 46, 50 and 54) for the same phosphorus levels. In all cases this may be seen to be chiefly a result of an increased number of high phosphorus moulds failing at less than 30 lives (Tables 47, 51 and 55).

The NEU 100 moulds, however, showed little apparent decrease in life at high phosphorus levels and there was a slight opposite trend but the number of moulds was too small to make any firm conclusions (Table 58).

It has been noted above (Section 2.5.4) that previous experience has claimed a beneficial effect of increased phosphorus levels[19]. This clearly is not the case for the Renishaw moulds manufactured with a high phosphorus burden, particularly for the F-type mould.

Banks[20] has reported that the presence of up to 0.4% phosphorus in ingot moulds is not deleterious, but great emphasis was placed on this being true only if the moulds were slow cooled after casting. However, mould temperature measurements have been taken at Renishaw immediately after stripping the top plate (Tables 60, 61 and 62), and these have demonstrated that for the F-type mould, when cast between 10.00 and 13.25 hrs. and stripped between 23.00 and 24.00 hrs., the surface temperature may lie between 890 and 960°C. Corresponding temperatures for the 610 OT, B120 and WEU 100 moulds were 610 to 780°C, 700 to 750°C and 500 to 770°C respectively. Drawings of the five mould types are shown in Appendix 2, and it is apparent that the much higher surface temperature for the F-type mould is due to its heavy base section (top as cast) compared with the other mould types.

It is clear then that the necessary slow cooling conditions for high phosphorus F-type moulds at Renishaw were not satisfied by the then current cooling practice. The combination of circumstances at Renishaw at the end of 1974 - high phosphorus levels, change of shift pattern, which resulted in the premature stripping of the F-type mould - undoubtedly produced highly crack-sensitive material, which manifested itself in the outbreak of premature failures at Tinsley Park Steelworks.

The cooling conditions and their relationship to the high phosphorus analyses for the 610 OT, B120 and WEU 100 moulds at Renishaw may be considered to be borderline and it is thought that this offers an explanation for their relatively small decrease in life compared to the F-type

mould.

The NEU 100 mould has a very light section at the top so that, although temperature measurements are not available, it may be surmised that it will be substantially cooler when the top plate is stripped.

The top temperature after removal of the plate is considered to be of greater significance than the removal of the sand jacket, since those F-type moulds cast on Friday and stripped within 24 hrs showed no decrease in life over those cooled for considerably longer (Table 43).

5.3 TITANIUM

5.3.1 Microstructure of Test Block Material

The microstructures found in the light section castings are consistent with those reported in the literature[29], i.e. fine undercooled graphite at high titanium levels and with a marked microstructural influence of gassing treatment a 0.2% addition of titanium being sufficient to produce Type D graphite with carbon dioxide treatment. This trend, however, is not as marked in the heavy section test blocks. In ingot mould sections, it may be surmised that at high titanium levels of up to 0.2%, under-cooled graphite will not be formed unless through some local segregation such as at the mould's upper surface.

The extremely coarse graphite structures, observed in Fullwood ingot moulds at high titanium levels, have not been produced in the experimental work. It is suggested that this is due to the lower carbon equivalent in the test blocks and to the much heavier sections of Fullwood moulds.

5.3.2 Microstructural Mechanism

There are three microstructural features concerning titanium additions worthy of further comment; namely, an initial coarsening of the graphite followed by under-cooling at higher titanium levels (i.e. 0.2% Ti), and the under-cooling effect of carbon dioxide and coarsening in hydrogen. It should be noted that the coarsening followed by under-cooling effect of titanium is similar to silicon, which also shows a coarsening effect on the graphite, through an increase in carbon equivalent. This is followed by the formation of under-cooled graphite at levels of 5% or above in the heat resisting Silal irons.

Liquidus concentrations for the FeCTi system are shown in Figure 69[48] and it may be seen that titanium also increases the carbon equivalent. Interpolating from the graph, it may be noted that 0.8% titanium decreases the eutectic carbon concentration from 4.3 to 3.9%. The carbon equivalent for the FeCTi system, at levels up to 0.8% titanium, may, therefore, be expressed by the formula

$$CEV_{FeCTi} = C + \frac{Ti}{2}$$

This should be compared with the more usual form of carbon equivalent value in the FeCSi system

$$CEV_{FeCSi} = C + \frac{Si}{3}$$

from which it may be observed that titanium in direct combination with carbon and iron is a more powerful graphitiser than silicon. It is suggested that the graphite coarsening action of titanium is due partly to its effect on carbon equivalent in a similar manner to silicon.

The action of titanium in lowering carbon equivalent

value in the FeCSiTi system is unclear but based on the above ternary systems and Comstock's observations[26] that the graphitising power of titanium is up to seven times that of silicon, the CEV may be expressed as lying between the two equations:-

$$CEV_{FeCSiTi} = C + \frac{Si}{3} + \frac{7Ti}{3}$$

$$CEV_{FeCSiTi} = C + \frac{Si}{3} + \frac{Ti}{2}$$

Olen and Heine have shown that for eutectic solidification to take place, austenite must first nucleate before eutectic graphite will precipitate out of solution [49]. This phenomenon has been observed by Heine and Loper[50] in cooling curve studies, where an initiating eutectic arrest occured in near-eutectic and hypereutectic irons at a slightly higher temperature than the main eutectic arrest. Henschel, Du Pont and Heine[51], in examining the effects of oxygen on the solidification of cast iron, noted that low oxygen potential atmospheres increased the initiating eutectic arrest effect, which was correlated with an apparent ease of nucleation and growth of austenite, whereas high oxygen potential atmospheres inhibited the nucleation of austenite and, hence, eutectic nucleation, causing under-cooling.

This mechanism may offer a possible explanation of the effects of carbon dioxide and hydrogen on cast iron microstructures. Bubbling carbon dioxide through the melt may be expected to increase the oxygen potential of the melt, thus inhibiting eutectic nucleation by its effect on austenite nucleation. The overall result will be an increased tendency to formunder-cooled graphite, which has

been observed in the test blocks. Similarly, the passage of hydrogen through the melt will reduce the oxygen potential, allowing freer eutectic nucleation and hence, a coarsening of the graphite.

It is interesting to note that both titanium and silicon are deoxidisers and part of the free energy diagram for oxides is shown in Figure 70. It may be expected that the addition of silicon and titanium to an iron melt will reduce the oxygen potential and increase eutectic nucleation thus coarsening the graphite, as well as this effect through increases in carbon equivalent, noted above.

At this juncture it may be worth considering the effects of aluminium. It has been shown that a 75% FeSi inoculant is more effective if it contains a small amount of aluminium, usually in the range 1-3%. This has been shown to improve the action of the inoculant in reducing chill and under-cooling. Similar effects have been noted for barium, strontium and zirconium. These, together with aluminium, are all powerful deoxidisers. The positions of aluminium, silicon and titanium in the free energy diagram are shown in Figure 70. It is tentatively suggested that their action may be identical with the mechanism noted above, i.e. increasing eutectic nucleation by their effect on oxygen activity in the melt.

The formation of under-cooled graphite at very high titanium and silicon levels appears to contradict Henschel, et al's hypothesis, since it might be expected that massive additions of titanium and silicon would reduce the oxygen activity to virtually zero. This would allow an easy austenite nucleation initiation, followed by eutectic

growth, to produce a well nucleated melt with little under-cooling. Clearly the under-cooling effect is worthy of further examination.

Oxygen concentrations in iron melts have been shown to be similar to nitrogen in the range 0.002 to 0.005% in induction melted iron[52]. It may be shown that the stoichiometric amounts of titanium and silicon to fix 0.005% oxygen as TiO_2 and SiO_2 are 0.0075% and 0.0044% respectively. This would indicate that only small additions of titanium and silicon are necessary to remove oxygen from solution whereas in practice relatively massive additions are required to produce under-cooling so that another mechanism other than oxygen removal, is in operation.

It was noted above (Section 2.5.6) that the formation of under-cooled graphite at high titanium levels has been attributed to its formation of TiS, thus reducing the sulphur concentration. However, this mechanism cannot be applied to the formation of under-cooled graphite at high silicon levels, since SiS shows no tendency to form in cast iron. It is considered that any explanation of the under-cooling effect should be able to include both titanium and silicon, because of their similar effects on microstructure, in particular their effects on graphite and pearlite.

Henschel, et al[51], in investigating the effect of carbon equivalent on bulk arrest temperature, opined that the degree of under-cooling may be related to thermal energy requirements or austenite surface requirements for the start of the eutectic reaction. Both titanium and silicon reduce austenite stability, so it is suggested

that, at high titanium or silicon levels, either the initiating austenite nucleation is inhibited or its surface is poisoned, so that eutectic nucleation is impaired, resulting in under-cooling.

5.3.3 Mechanical Properties of Test Block Material

Reference to Figures 14-17 shows that there is an increased strength and embrittling effect over the approximate range 0.02 to 0.04% titanium, for the hydrogen and untreated, heavy section test blocks, although this embrittlement occurs at a slightly higher titanium level in the carbon dioxide treated test blocks. This effect was also observed in fracture toughness tests on ferritic irons[41], although a minimum in the tensile strength was observed. It was postulated that the reduction in fracture properties was due to the removal of nitrogen from solution causing a weakening of the ferrite. The pearlitic irons under current investigation are anomalous with respect to this hypothesis, since they contain little ferrite (although the small amount present is associated with the graphite flakes) and over the critical range under consideration an increase in tensile strength is observed. No metallographic features were observed over this titanium range (0.02-0.04%) to offer an explanation for the embrittlement.

Maximum ductility in the heavy section test blocks is developed over the range 0.05-0.12% titanium (Figures 15-17). However, this region also corresponds to the minimum tensile strengths (Figure 14). It would be expected that, although ingot mould cracking resistance would be improved over this region, any improvement in

life would be offset by reductions caused by broken lugs, etc.

At the highest titanium levels in the heavy section test blocks, brittleness again occurs. This is attributed to the formation of some under-cooled graphite in the microstructures. As discussed above, this is unlikely to occur in ingot mould sections, so that the improvement in ductility may be expected to continue to higher titanium levels, with a corresponding decrease in tensile strength, than that indicated in the present work.

5.3.4 The Effect of Titanium on Ingot Mould Performance

5.3.4.1 Fullwood Slab Moulds

Moulds Produced from 70 Pig, 20 Steel, 10 Returns Burden

During the period from mid 1974 to mid 1975, Fullwood produced a total of 155 48-type moulds from a 70% pig iron, 20% steel and 10% return scrap cupola charge. The mean carbon, silicon, manganese and titanium levels of these moulds were 3.96%, 1.52%, 0.75% and 0.039% respectively with a mean carbon equivalent of 4.47%. Mould microstructures consisted of coarse graphite (mean USV = 3.54 km/s) in a predominantly pearlitic matrix. Unfortuneately it was only possible to obtain bend properties from mould No.585 after failure at 120 lives, and so the bend properties (A/B = 0.6) are not characteristic of the as-cast properties.

The performance of the 48-type moulds produced from the 70/20/10 burden was satisfactory with a mean life of 67.88 lives. The level of repairs on moulds from this burden was high (51%, Table 29) but it is impossible to

know how this affected the final performances. The majority of failures were due to vertical cracking, although a significant proportion of the failures (10.1%) were reported as being due to horizontal cracking. It is probable that a high proportion of the horizontal cracks were, in fact, due to cold shut defects, in view of the low mean casting temperature (1237°C).

Moulds produced from 53 Pig,

17 Steel, 20 Returns, 10 Broken Mould Burden.

The moulds produced from the 53% pig, 17% steel, 20% returns, 10% broken moulds burden, had mean carbon, silicon, manganese and titanium levels of 3.91%, 1.47%, 0.75%, 0.03% respectively, with a mean carbon equivalent value of 4.4. The mould microstructures were varied, and, metallographically, it was not possible to identify a characteristic graphite morphology. The mean USV value (3.65 km/s) indicated a slight refinement of the graphite consistent with the lower mean carbon, silicon and titanium values. Bend properties indicated a slight embrittlement (A/B 1.2-3.7), although it must be appreciated that these samples were obtained from two low life moulds, and are not, necessarily, characteristic of the group of moulds.

However, the reduced CEV is consistent with an iron of lower ductility, which, in turn, agrees with the impaired performance of moulds produced from this burden. The most significant feature of the early failures experienced on moulds produced from the 53/17/20/10 burden, however, was the severity of the premature cracking, and it was this feature, above all others, which lead to the decision to alter the burden, in June 1976 (see fig. 33).

In terms of overall performance (Table 27 and fig.32), the performance of moulds produced from the broken mould burden were not as disastrous as was feared, at first, although it is clear that, statistically, the moulds gave an inferior performance to moulds produced from the earlier (70/20/10) burden.

The impaired performance of the moulds produced from the 53/17/20/10 burden was, therefore, due to the lower carbon, silicon and titanium levels, which produced a more crack-sensitive iron. This was reflected by a marked increase in vertical cracking (75.5%), especially on the narrow side. The increase in cold shut defects (17.3%) is not due to metallurgical features, but probably reflects the low casting temperatures (mean value of 1236°C) experienced in this period.

Moulds produced from 63 Pig,
17 Steel, 20 Returns Burden.

The decision, during June 1976, to change the burden to 63% pig/17% steel/20% returns was taken to eliminate the broken mould additions, which seemed to be related to the problem experienced with the 53/17/20/10 burden. The first five moulds failed by cracking with an average life of about thirty (Fig. 35) and it was decided, therefore, in November 1976 (after 67 Type 48 moulds had been produced) to add titanium to the cupola, since previous experience had shown controlled levels of titanium (0.05% Ti) to be beneficial.

Analysis of the complete performance data on the 63/17/20 burden, however, shows a satisfactory mean performance of 63.34 (Fig. 32). The mean carbon, silicon,

manganese and titanium levels of the moulds were 3.92%, 1.49%, 0.76% and 0.04% respectively, with a mean carbon equivalent of 4.42%. Variable graphite microstructures were observed, although the mean USV of 3.58 km/s suggested an increase in graphite size over the 53/17/20/10 burden. Inadequate bend test data was available for this burden to be able to determine the characteristic material properties.

However,a significant improvement in mould performance was achieved over the 53/17/20/10 burden, and this is due to the increased titanium content which, it is suggested, reduced the strength, improved the ductility and thus improved the cracking resistance in service.

The beneficial effect of replacing the broken moulds by pig iron at Fullwood may be explained in two ways. Firstly, analysis differences were caused because mould scrap has a lower carbon content, typically 3.8%, as opposed to 4.0-4.2% in pig iron, lower silicon (1.5% against 1.5-2.0% silicon specified for the pig iron) and a lower titanium content. Titanium contents in Workington pig iron for the 1.5-2.0% silicon grade are typically 0.055% , mould scrap at the time being 0.04%. Secondly, the physical size of broken mould scrap is much larger than pig iron. It has been shown[53] that larger scrap melts further down the bed allowing less superheat to occur. Simultaneously, carbon pick up will be less. During cupola operation at Fullwood, pieces of broken mould have been observed at the level of the cupola spy hole.

It is interesting to note that in spite of the improved mould performance over the 53/17/20/10 burden, the

proportion of moulds failing by vertical cracking increased to 82.1% (Table 30) but the incidence of narrow side cracking decreased. Although the relative proportions of repaired moulds were the same in each burden (Table 29), the cracking in moulds from the 53/17/20/10 burden was more serious in that it tended to occur earlier in life, thus resulting in a lower mean life.

A significant proportion (11,9%) of moulds from the 63/17/20 burden had to be broken by drop balling (Table 30) to remove sticker ingots. This supports the view that the cracking resistance of moulds from this burden was improved.

Moulds Produced from 63 Pig Iron,
17 Steel, 20 Returns + FeTi Burden

The reasons for adding ferro-titanium to the 63/17/20 cupola burden charge in November 1976, were given earlier. The mean carbon, silicon, manganese and titanium levels in the moulds were 3.92%, 1.50%, 0.73% and 0.051% respectively with a mean carbon equivalent of 4.42% (Table 32). Coarse graphite stuctures (mean USV = 3.46 km/s) in a predominantly pearlitic matrix were produced, and were due mainly to the increased titanium levels.

The mould performance from this burden (63 moulds failed) maintained an acceptable level of 65.03 lives. Three of the first five moulds from this burden to fail, in fact had less than five lives each, but subsequent improved performance proved this was simply due to the normal patterns of failure.

The improved performance of the titanium treated moulds may be attributed to the factors examined for the 63/17/20 burden, the higher titanium levels producing a

weaker iron with better resistance to cracking (this despite the observation that 86.3% of these moulds failed by vertical cracking). However, the incidence of torn seat defects increased fron nil (63/17/20) to 5.9%, with the addition of titanium, and it is clear that the titanium level should not be allowed to exceed the 0.06% level.

Effects of Analysis Changes on Performance of Fullwood Moulds at Ravenscraig

Analysis of all ingot mould production data, has now provided an understanding of the changes in performance of the 48-Type moulds with changes in cupola practice, and this has been reinforced by the regression equations. It is clear that the performance of the 48-Type at Ravenscraig is critically dependent on the mould analysis, which affects the cracking resistance. The severe deterioration in performance which occurred with the introduction of broken moulds into the burden was due to the reduction in carbon equivalent, which produced iron of reduced cracking resistance. Subsequent burden changes increased the graphite coarseness by either simply increasing the carbon equivalent or by increasing the titanium content.

This effect of titanium is shown graphically in Figure 71, which includes data for all moulds manufactured after 1972, with controlled stripping. It may be noted that, as titanium is increased over the range 0.02 to 0.10%, there is a decrease in the incidence of vertical cracking and an increase in crazing and torn seats. This behaviour correlates with the microstructural examination of failed moulds which shows a progressive increase in graphite size over this titanium range, resulting in increased toughness

(i.e. reduced cracking tendency) but reduced strength (i.e. increased tendency to fail by broken lugs, torn seats, etc.). The overall effect on mould performance is to give a maximum in the titanium range 0.04-0.07%.

In addition to the improved performance brought about by increased Ti levels, it is interesting to note that the addition of broken moulds to the burden caused a reduction in Ti from 0.04% down to 0.03%. It is possible that this (0.03%) Ti level corresponds to the ductility trough for cupola iron. Unfortunately, there is little experimental data in this area, so that further work is required to support this argument.

5.3.4.2 River Don Forging Moulds.

An important factor in the reasons for the plucking of the River Don moulds is considered to be the low strength of the iron. Ultrasonic velocities were measured on Rheinstahl mould 105/101 No. 777 and found to be 3.7 km/s compared with 3.3 to 3.5 km/s for the River Don moulds. Typical analysis of Rheinstahl moulds are:

C	Si	S	P	Mn	Ti
3.7-3.9	1.5	0.04	0.04	0.7	0.01-0.02

and it is noticeable that their practical failure mode is one of cracking rather than plucking. It may be seen, therefore, that the difference in ultrasonic velocities may be a significant factor in the plucking of the River Don moulds and results from their higher Si(up to 2.0%) and Ti (up to 0.065%) contents. This is supported by the coarse graphite microstructures and the very ductile bend test results. Efforts have been made to limit Si and Ti in later moulds, but it is not yet known if this action

has had a significant effect on performance. Particular problems have occurred in obtaining low Ti levels. The principal source of Ti is from borings, but it is necessary to have a significant proportion of this material in the charge for satisfactory furnace operation. Although low Ti ingot mould scrap may give the required Ti analysis, problems have been experienced in melting at an economic rate.

Any increase in strength will, necessarily, result in a decrease in ductility, so that if Si and Ti levels are reduced to too low a level, major cracking may result. Two moulds produced by River Don have cracked, both containing high Ti levels and low ultrasonic velocities. Both moulds were repaired by stitching and satisfactorily reused.

It has been suggested that the cracking of Mould 105/101 No. 4 after first usage may have been caused by high residual stresses. The procedure of stress relieving of small moulds has been shown to be of doubtful value[54] but pre-warming of large forging moulds, prior to the first cast, has been claimed to be beneficial[55]. Whether this beneficial effect occurs through a reduction in residual stresses or a reduction in stresses induced by thermal shock, is uncertain. The low temperature used, of 200°C, suggests, however, that the latter mechanism is more likely to be operative. As more information becomes available, it may become necessary to review the possibilities of a pre-heating/stress relieving cycle being applied to River Don moulds.

Not forgoing the above consideration of the effects of weak iron in contributing to surface plucking, it should be remembered that the occurence of a material of low thermal conductivity, at a small distance behind the hot face, will also have a major effect. This causes increased disruption of the surface, because of relative differences in thermal expansion, and the interface will act as a line of weakness, to aid plucking. Such non-metallic inclusions were observed in the River Don moulds and it should be emphasised that steps should be taken to avoid their entrapment.

5.3.5 Use of Mould Scrap in Electric Furnaces

For convenience, the A/B bend test parameters for the mould scrap test blocks are plotted in Figure 72 together with results from the other heavy section test blocks. It may be seen that the results for the clean scrap closely fit the results for test blocks cast without any gassing treatment. The dirty mould scrap properties are similar to the carbon dioxide treated test blocks at 0.06% titanium and worse at 0.037% titanium although there were no carbon dioxide treated blocks at this level. It may be expected, therefore, that where electric melting is used based on a charge of dirty, oxidised scrap then the effect of titanium will be as if the iron had been treated with carbon dioxide, that is a ductility trough will be experienced at around 0.03 to 0.07%.

In the case of cupola melted iron conditions may be expected to be reducing so that titanium in iron produced by this route will behave in a similar manner to the hydrogen treated test blocks. A/B parameters measured on

test blocks and ingot moulds together with the hydrogen treated heavy section test blocks are plotted against titanium content in Figure 73. The cupola melts agree well with the experimental melts and confirm the existance of a ductility trough at 0.035 to 0.045% titanium.

In the case of cupola melted iron it was noted in Section 5.3.4.1 that optimum ingot mould performance for Fullwood 48-Type moulds occurs at 0.04 to 0.06% titanium. It is apparent, therefore, that should Fullwood convert to electric melting using dirty scrap this titanium range would now fall into the ductility trough which might result in a reduction in mould performance. In order to avoid this it would be necessary to increase the titanium content into the range 0.08 to 0.10%.

It is clear, therefore, that in considering an optimum titanium analysis to control mould cracking it is necessary to consider the type of furnace to be utilised and also the scrap condition.

5.4 NITROGEN

It was noted in Section 2.5.6, that during the 1960's, hot blast cupolas were introduced within the ingot mould foundries, thus enabling larger amounts of steel scrap to be used in the burden. Performance of moulds at that time was, in some cases, poor and examination of moulds revealed the presence of compacted graphite. Since the formation of compacted graphite was attributed to high nitrogen levels, the higher nitrogen levels found in these irons were blamed for the poor performance. To effect a remedy titanium additions were made, to fix the nitrogen as Ti(CN) and

performance improved.

The metallographic examination of the Fullwood slab moulds demonstrated several important points in connection with nitrogen and graphite compaction. Firstly, that the presence of compaction was unrelated to nitrogen content, and, secondly, that the presence of compacted graphite was not, in itself, detrimental to mould performance. This latter view is also supported in that calcium cyanamide treated ingot moulds have given satisfactory performances. The performance of small, square moulds has also been shown to be unrelated to graphite compaction[56] and the same investigation showed no correlation between high nitrogen levels and graphite compaction. Fracture toughness tests have shown that compacted graphite iron is tougher than flake iron[57].

One of the main difficulties associated with the metallographic examination of failed ingot moulds is that examination is, generally, confined to premature failures. Consequently, the reasons for failure are often attributed to certain metallographic features, which, however, are also present in good quality moulds. Furthermore, mould microstructures are so heterogeneous that extremely wide variations in mould microstructure may be observed in samples from the same mould. For example, moulds have been examined which contain flake graphite on the outside faces and compacted graphite in the centre[57]. From the examination of Fullwood slab moulds it is now established that compacted graphite occurs in good performance moulds, and it is not an uncommon feature of Distington moulds. Its appearance in Distington Mould 758, for example, was not

associated with premature cracking (Section 4.3.1.5).

The use of metallographic examination should not be used in isolation to decide on causes of mould failure. However, metallographic examination should still be regarded as an important part of a mould failure investigation, but it is only in very rare cases when such examination is likely to provide the solution. This was illustrated in the Fullwood slab moulds, in the case of mould 774 (Section 4.3.1.5) which consisted of a fine pearlite matrix, due to chromium contamination. All the other prematurely failed moulds contained such variable microstructures that it was impossible to attribute failure to a common metallographic feature.

The experimental test block material with deliberate additions of nitrogen has confirmed that high nitrogen levels, per se, do not cause graphite compaction. This may be seen by comparing the microstructures of casts RG666A and RG752A, with 0.0054 and 0.018% nitrogen respectively. Reference to Table 20 also shows that, as nitrogen levels are increased, there is no accompanying decrease in cracking resistance.

So called "nitrogen compacted" flake iron has a similar graphite microstructure to that in quasi-flake iron as revealed by deep etching, although the surface of the flakes are more irregular in the former. Quasi-flake iron itself has been shown to be an excellent ingot mould material giving improved performances over conventional flake graphite moulds (although this is only usually found where the matrix is ferritic). For a given matrix microstructure it appears logical to conclude that moulds with

compacted graphite will be similar to quasi-flake graphite moulds. The presence of compacted graphite, therefore, may not, in itself, be detrimental to mould performance.

Despite the foregoing comments, one of the reasons why the nitrogen compaction theory of mould embrittlement has remained in favour is that corrective measures to remove nitrogen have generally produced a marked improvement in performance. Titanium, for example, reduces the cracking resistance of cast iron, and its effect in producing coarse graphite is usually cited as a result of removing nitrogen as Ti(CN).

Nitrogen will also act as a pearlite stabiliser and solid solution strengthener and may also affect mould performance by these mechanisms. It has been shown that as nitrogen is removed from solid solution by titanium additions, there is a marked reduction in cracking resistance in ferritic irons[59], so in these irons, the presence of nitrogen may be beneficial. In pearlitic irons there is a similar reduction in cracking resistance but the precise mechanism has not yet been determined.

Reasons for the formation of compacted graphite have not yet been resolved by the present work, but a few observations may be made. Earlier work, by other authors, showed traces of compaction when additions of nitrogen bearing salts were made to the melt, calcium cyanamide being most favoured[38]. As noted above, other elements are also introduced with the addition, for example calcium and sodium. The present work has shown that the amount of calcium pick-up by the melt is small (Table 19) and that, even at high calcium levels, no direct evidence of

compaction was observed (Figure 25). Calcium, however, is used as a graphite modifier, particularly in Japan. There is little information available on the effects of sodium in cast iron, but it has been shown to have similar effects to Ba[59] (which is in the same period as Ca). Sodium, therefore, might be expected to be capable of graphite modification.

It has long been known that compacted graphite is found only in heavy section iron typical of ingot moulds, so that ferrostatic pressure and/or cooling rate may contribute to its formation. A recent examination, however, showed that the mechanism of compacted graphite formation is more complex[57]. Samples were taken through the wall of a mould, produced at Stanton, and the inside and outside faces were found to contain flake graphite, but the centre portion was compacted. There are several possible explanations of this behavior. Firstly, there may be a critical cooling rate for the formation of compacted graphite; however, identical moulds may, or may not contain compacted graphite, so that cooling rate may not be the sole criterion. Secondly, segregation may be important. It is well known that, even in nominally flake irons, graphite particles in the inter-cell boundary regions may show evidence of compaction. Since this is the last metal to solidify, it is likely that segregation effects are causing enrichment of certain elements in the boundary area, which promote compaction.

On a macro scale, however, because of the nature of iron solidification, it is considered that segregation is likely to play a major role in the effect observed above.

Thirdly, ferrostatic pressure has been suggested as a possible cause of compaction. Since the ferrostatic pressure across the mould wall will be uniform, it may be concluded that this is not the sole cause of compaction.

In addition to ferrostatic pressure, considerable pressure is exerted during solidification, by the volume changes associated with graphite growth. The precise effects of these forces will depend, to a large extent, on the degree of mould dilation. With soft moulds, such forces may be expected to be relieved.

An electron metallographic examination of flake and compacted graphite irons has been carried out (see section 4.3.1.6.) and showed that flake irons contained a distribution of fine and coarse TiN precipitations of 0.03 μm, and 0.15 μm diameter, respectively. Compacted irons, however, tended to contain less (or none) of the finer particles. The precise significance of this in the mechanism of compacted graphite formation is unclear. The absence of TiN particles may indicate, either that there is a lower titanium level in compacted irons, or that the nitrogen levels in solution are higher. At present, the use of soluble and insoluble nitrogen analysis figures refer to their solubility in the chemical solutions used for extraction. A method of analysis to determine "mobile" nitrogen levels by a heating technique, is under current development. Until this technique is better established, it is considered that the roles of nitrogen in solution or as precipitates may not be evaluated.

The presence of high residual levels has also been found to enhance graphite compaction and this is discussed

more fully in Section 5.5.

It may be seen, therefore, that the formation of compacted graphite is complex and the following factors have been identified as being contributory:-

a) Nitrogen,

b) Presence of graphite modifiers, e.g. Ca and Na,

c) Cooling rate,

d) Ferrostatic pressure,

e) Trace element levels.

The occurrence of porosity in cast RG 752A should be recognised as a potential problem. Such porosity will alter the thermal properties in the surface region and is likely to cause increased mould wall temperatures and, hence, stresses and the mechanical strength will be reduced. Such porosity should, therefore, be avoided. Under the conditions employed in the present experiments porosity occurred at 0.018% nitrogen, but none was found at levels of 0.012 or 0.014%.

It may be seen, therefore, that the principal effects of nitrogen on ingot mould performance will be through its action as a pearlite stabiliser, and through the formation of gas porosity, in which case low levels of nitrogen are desirable to prevent cracking. In either case the presence of nitrogen levels as high as 0.012% have not been shown to be detrimental. Whether nitrogen is contributory in forming compacted graphite or not, this morphology is not considered to be necessarily detrimental.

Several methods of increasing melt nitrogen levels have been tried in the present work. Nitrogen bubbling did not give increased nitrogen levels, actually decreasing

them. This is considered to arise from the purging action promoting nitride particle removal and also allowing equilibrium to be achieved more rapidly. The pick-up of nitrogen in cupolas, therefore, would appear to be difficult but it is considered to occur by nitrogen dissolving in local areas of low carbon equivalent; for example, from a piece of melting steel. Use of nitrogen or air in porous plug treatments is also known to give little nitrogen pick-up.

Of the other two methods of increasing nitrogen levels it is considered that the use of high nitrogen ferromanganese is most effective. The use of calcium cyanamide is fairly efficient but it generates considerable fume and there is the problem of possible increase of calcium levels. It is considered, therefore, that NFeMn offers the best method of increasing melt nitrogen levels.

Pick-up of calcium from Calsiloy is effective and is suitable for experimental purposes. Care should be taken to compensate for the resultant increase in silicon levels.

5.5 RESIDUAL ELEMENTS

The importance of residual elements and their effect on ingot mould performance is demonstrated by the examination of prematurely failed Round Oak WEU100 moulds. Regression analysis of the data for this mould type has shown that:-

$$\text{Life} = 82.89 - 320.1 \times (\%P) - 2591 \times (\%Sn) + 299.1 \times (\%Ti) + 269.7 \times (\%Ni),$$

and the mean analyses are shown in Table 63[60].

Analysis values for the failed moulds and the mean analysis have been substituted into this equation and the

results are given in Table 63. Mould number 489 shows a drop in life, due mainly to the high tin content. In general a drop in life of 10- 20 below the average life of 71 is predicted. However, chromium, which did not appear in the regression analysis because only a small variation was found in the statistical sample, would be expected to have a more serious effect than tin, so that the low performance of mould numbers 404 and 587 may be attributed to this effect. In fact massive carbides were seen in mould number 587 (Figure 64) to support this view.

The regression analysis indicates that the WEU100 mould would give an improved performance with a partially ferritic matrix. This could be achieved by a reduction in Mn, Cr, Sn, Mo and Cu levels with an increase in Si and Ti. The microstructures of the failed moulds were fully pearlitic.

The bend test results show a marked difference between samples taken from the base and 150mm up the wall for mould number 404. This is shown in Figure 59, which are typical bend test curves from these positions and demonstrates that adjacent to the base the material is more brittle. The microstructure in this region consisted of fine pearlite, compacted graphite and numerous MnS inclusions (Figure 60b) whilst 150mm from the base, the structure consisted of flake graphite, fine pearlite and some carbide/phosphide eutectic (Figure 60a). These observations may be due to either segregation or to premature stripping of the top plate.

An examination carried out at Renishaw (see Section 4.2.3.4) indicated that there was little control over

removal of the top plates. The maximum temperature recorded for the WEU100 moulds was 770°C, 13¾ hours after casting, but it seems likely that if moulds were cast at the end of the day this could be significantly higher.

In general, all the bend test properties were typical of brittle iron.

The effect of a high chromium level in Fullwood slab mould number 774 has already been mentioned.

Examination of the microstructures from the test block material, shown in Figures 28 to 30, demonstrates that increased residual content and reduced cooling rates increase the amounts of intercellular carbide and compacted graphite. Point counts, to determine the volume fraction of carbide, were carried out on four samples and the following results obtained:-

RG720A (fast cooling, low residuals) - 0.3%

RG867A (fast cooling, high residuals)- 0.5%

RG852 (slow cooling, low residuals) - 0.6%

RG868 (slow cooling, high residuals)- 0.9%.

It may be seen that the higher residual levels increase the amount of carbide as might be expected, but that this effect is apparently increased by slower cooling rates. This is seen to occur since slower advance of the solidification front allows greater time for segregation to occur.

The resultant mechanical properties in Figure 27 show that, in high residual melts, the more slowly cooled test blocks are more brittle and this is considered to arise from the increased amounts of inter-cellular carbide. In the low residual melts, the carbide present shows a much

lower tendency to segregate to the cell boundaries, so that this results in improved ductility. In addition, the ductility of the slow-cooled blocks is higher than the fast-cooled blocks, i.e., in the opposite manner to the high residual melts. This anomaly may be explained by consideration of the lower volume fraction of carbide and its less segregated nature being less significant in the low residual melts. It is considered that the major contribution to ductility variations in the low residual melts will be traces of ferrite surrounding the graphite flakes and the pearlite interlamellar spacing, and the overall graphite coarseness all of which will be worsened by faster cooling rates. In addition, irrespective of cooling rates over the range examined, increased residual levels cause a marked increase in brittleness.

The application of the above results will depend, ultimately, on ingot mould size. For large slab moulds, cooling rates are extremely slow. Trials at Fullwood Foundry[61] have shown the time to cool to 700°C for the 25t 48-type mould is of the order of 60-70 hours. The slowest time in the experimental material is appreciably faster, at around 15 hours (Fig. 16). It may be noted, therefore, that the observed detrimental effects of reduced cooling rates will be appreciably magnified in large slab moulds. Since the principal failure mode of slab moulds is one of cracking, it is apparent that reduced residual levels will offer significant improvements in cracking resistance and, hence, ingot mould performance.

It was seen above (section 2.5.8.), that the Distington residual levels are lower, generally, than

those of the other foundries and this may, to a large extent, explain their generally superior slab mould performance. It is considered, therefore, that a general reduction in trace element levels to those employed at Distington, or even lower, would improve slab mould performance. It should be remembered, however, that the use of lower residual materials will increase furnace charge costs, so that an analysis of the cost benefits of improved slab mould performance versus increased charge costs, is required.

In small, square moulds, the cooling rates are appreably faster, with times down to 700°C in approximately twelve hours. These times are, nonetheless, slow compared with the smaller test blocks used in the present investigation. It may be expected, therefore, that some improvement in cracking resistance would be obtained by reducing residual levels in small, square moulds, although the potential improvements are relatively lower compared with slab moulds.

The issue is complicated, however, in that small, square moulds, in general, fail by crazing, so that, although cracking resistance is important, improvements in performance will be obtained by improving crazing resistance. It is generally recognised that this may be achieved by increasing the stability of the pearlite. Since the residual elements considered in this investigation increase pearlite stability, it may be seen that, where moulds are failing by crazing, there is no advantage to be gained by reducing residual contents below their current levels.

It is considered that an adjustment of the maximum

trace element levels quoted in the Code of Practice for large slab moulds may be of benefit. Suggested levels are as follows:-

Sn	Cu	Ni	As	Cr	Mo
0.015 max.	0.01 max.	0.05 max.	0.01 max.	0.05 max.	0.01 max.

It should be remembered that these are maximum levels so that mean analyses will be appreciably lower. It is considered that the current levels are suitable for moulds failing by crazing.

5.6 COOLING RATE AND STRIPPING.

The effects of fast cooling from temperatures of around 1000°C by removal of all or part of the sand jacket has clearly been shown to be detrimental to ingot mould performance with respect to phosphorus (Section 5.2) and to residual levels in the case of WEU100 moulds (Section 5.5). Even at low phosphorus levels the test blocks show a marked increase in brittleness with early stripping (Table 9). This was also confirmed at high and low nitrogen levels (see casts RG667 and RG668, Table 20). Clearly, then, the whole sand jacket must remain in position until the casting has fully transformed, that is until, say, 650°C.

With the sand jacket remaining in position until fully transformed the effects of variations in cooling rate have been shown to depend upon residual levels (see Section 5.5). In practice this cooling rate is adjusted by controlling the sand thickness and is normally around 125 to 150mm although this may be considerably thicker in the case of slab moulds or where moulds of varying sizes are produced in the same box size.

The introduction of furane sand systems, for ingot mould production at Distington and Fullwood, has caused interest in the possibility of reducing sand thicknesses, so as to reduce production costs[61]. For slab mould production at Fullwood and Distington a reduction in sand thickness from 300mm down to 150mm would bring about a reduction in cooling time, down to 700°C, from 65h to 45h, this latter thickness and time being in line with ingot mould practice at Hoogovens, Holland. Since cooling times of the test blocks used in the present experiments are much faster than these rates, the likely effect on properties of low residual melts may not be stated. The use of 355mm sleeved test blocks gives the slowest cooling rate possible at Sheffield Laboratories. It is considered, therefore, that effects of sand thickness at these slow cooling rates require to be investigated by work's trials, sufficient moulds being produced with a reduced sand thickness to detect a statistically significant difference in mould performance.

5.7 CASTING TEMPERATURE IN EXPERIMENTAL CASTINGS AND INGOT MOULDS

No difference in microstructure in the test blocks was observed, all being flake graphite with a trace of compacted graphite in the cell boundaries. Pearlite contents were 85 to 95%. The pearlite interlamellar spacing (Table 24) showed no systematic variation. The most significant structural variation was in the eutectic cell size, which was decreased by a reduction in casting temperature and by the addition of ferrosilicon, Table 24, due to its inoculating effect.

The ratios A/B and a/b, Table 24, show that the cracking resistance of the iron increases as the casting temperature was reduced from 1300 to 1200°C. A further increase in cracking resistance was observed with ferro-silicon addition.

The most notable correlation appears to be that of eutectic cell size with bend test parameter, and this is shown graphically in Figure 31. This decrease in eutectic cell size was observed by Banks[11]. Banks, however, demonstrated the formation of compacted graphite at a casting temperature of 1200°C which has not been observed in the present work.

The present results suggest that the cracking resistance could be improved by lowering the casting temperature below 1250°C. Further, the beneficial effects of an addition of ferrosilicon, prior to casting are marked.

It was noted in Section 5.3.4.1 that the incidence of cold shuts and horizontal cracks for Fullwood slab moulds was 11.9 to 18.3% for the 70/20/10 and 53/17/20/10 burdens where the casting temperatures were 1237°C and 1236°C respectively. In subsequent burdens the mean casting temperatures were increased to 1250/1255°C and the incidence of these failure modes was reduced to 2-4½%. It may be seen, therefore, that in slab moulds any potential benefit in improving cracking resistance by reducing casting temperature is likely to be outweighed by the increased incidence of cold shuts. 1250°C appears to represent the minimum acceptable casting temperature for this application.

The statistical examination of the 610 WB moulds (4.5t) showed that, as casting temperature is increased

from 1200°C up to 1260°C, there is a gradual increase in performance from around 70 usages up to 85-90 usages (Figure 67). At higher casting temperatures than this there appears to be relatively little effect. This beneficial effect of increased casting temperature is also reflected in the positive coefficient in the regression equation.

The effect of casting temperature on the relative amounts of cracking and crazing in the 610 WB moulds is shown in Figure 68. Appreciable scatter in the results is apparent, particularly at high and low temperatures. Since there are over 80% of moulds failing by crazing, if only those groups where there are over 100 moulds are considered i.e. the casting temperature range 1240 to 1309°C, some slight reduction in the incidence of crazing may be seen with increasing casting temperature. This indicates that reduced casting temperatures may increase cracking resistance, agreeing with the experimental test block findings. Reference to Tables 74 and 75 shows that where the mean casting temperature for crazed moulds is significantly different to that for other failure modes, the life is not significantly different. It may be surmised, therefore, that the effects on cracking resistance brought about by changes in casting temperature have little or no effect on mould performance in this mould type.

Because of this latter observation it may be surmised that increased casting temperatures improve the crazing resistance of the iron. It may be seen, therefore, that for small square moulds failing by crazing, principally, such as the 610 WB mould, that improved performance may be achieved by maintaining the casting temperature above 1270°C.

Mould type L26 was chosen since it was expected that a greater percentage of moulds might fail by cracking. Although this is the case the total number of moulds available for analysis is too small so that no clear trends of the effect of casting temperature may be discerned.

5.8 METAL SUPERHEAT

Reference to Table 26 shows that variations in metal superheat temperature in the experimental test blocks has had little effect on material ductility although a slight increase in tensile strength may be discerned with increasing temperature. Because of furnace design and operating practices it is not possible to hold iron at temperature for periods longer than the 15 minutes used in the present work. This time, clearly, is not representive of an electric furnace practice based on maintaining a large molten heel where residence times at temperature may regularly amount to hours or even days over holiday shutdowns. Examination of the effects of superheat temperature on this type of practice will, clearly, require a statistical evaluation of mould performance data. The present laboratory results, however, do indicate that brief periods at elevated temperature appear innocuous.

5.9 MATERIAL FROM INGOT MOULDS FOR ROUTINE ASSESMENT

Examination of Table 65 shows that the horngate samples taken at Renishaw have given unusually high a/b values compared with a/b ratios from lug samples (Tables 66 and 67). The horngates are approximately 50x100x200mm which is considerably thicker than the 25mm thick lug samples so that the cooling rate would be expected to be slower and hence give better ductility. In normal ingot

mould foundry practice, however, the top and bottom plates are removed and the mould, sand jacket, core and metal casing removed to the cooling area. Since the horngate is located on the bottom face this practice will expose the horngate at elevated temperature thus giving an abnormally fast cooling rate. In addition, the moulds after top and bottom plate removal are frequently placed on damp sand beds. This may be expected to further accelerate the cooling rate. If the sand jacket is removed from the mould then this will be reflected in the cooling rate and hence properties of the lug sample whereas the horngate sample will be unaffected.

For a routine test to determine the properties of an ingot mould it is necessary for the test piece to be representative of the mould as a whole. This is clearly not the case for the horngate samples. The large lug samples, therefore, appear best to fulfil this function.

For the Renishaw lug samples (Table 66) two moulds gave considerably higher a/b values than the remainder and one of these moulds failed by cracking, albeit at close to average life. For the Distington lug samples (Table 67) only one mould cracked and the a/b value for this mould was slightly lower than average. The performances of moulds 6813 and 6728 at 122 and 40 lives are noteworthy since both moulds possess high a/b ratios and both failed by crazing. It should be remembered, however, that the bend test is only capable of ranking the cracking susceptibility of ingot mould iron.

As residual levels increase crazing resistance increases and cracking resistance reduces. There must, therefore,

be an optimum level at which crazing is reduced to such an extent that a high life is obtained without the mould cracking. In terms of the bend test a/b parameter this means that a high value may indicate both a high life mould if it fails by crazing or one susceptible to cracking. In any study of this kind, however, the issue is further complicated by variations in usage pattern from mould to mould so that, by chance, a crack prone mould may be used in such a manner as to delay cracking and promote crazing.

Notwithstanding the above discussion, the sucess of the Renishaw lug samples is seen to indicate that, indeed, the large lug samples tested using the three point bend test are suitable for the routine monitoring of ingot mould properties. The parameter a/b used in this work is easy to measure and interpretation of the results is simple, requiring non-skilled operators.

The Distington lug samples represent the largest population of moulds from which bend test samples have been taken, and for this reason the regression analyses were carried out.

Examination of the correlation matrix shown in Table 69 shows that the bend test parameters maximum load, a/b, A and A/B all show high correlations with each other as might be expected. The parameter B only correlates, negatively, with A/B. It is noteworthy that carbon shows significant negative correlations with maximum load, a/b, $A_{\frac{1}{2}}/B_{\frac{1}{2}}$, A and A/B as do silicon and titanium, although these are less significant.

The only significant variable with life in the correlation matrix is silicon. In the regression equation phosphorus

$A_{\frac{1}{2}}/B_{\frac{1}{2}}$ and A/B also appear. It is interesting that the coefficients of A/B and $A_{\frac{1}{2}}/B_{\frac{1}{2}}$ are negative and positive, respectively. The numerical value of the A/B coefficient is five times larger than that for $A_{\frac{1}{2}}/B_{\frac{1}{2}}$ so this might be expected to have the more dominant effect. It may be seen, therefore, that life would be more affected by the A/B ratio, increasing A/B value reducing life as may be expected.

5.10 COMPARISON OF CHARACTERISTIC FOUNDRY MATERIAL

Examination of the bend test results given in Table 71 shows a marked difference in bend test properties for each of the foundries. In order of increasing a/b ratios the foundries may be ranked as follows:-

Distington (2.3)
Fullwood (2.5)
Dowlais (2.6)
Landore (3.6)
Craigneuk (5.4).

Recent typical mould performances for these foundries taken from quarterly mould summaries have been as follows:-

Distington (98 lives)
Fullwood (76 lives)
Dowlais (66 lives)
Landore (64 lives)

where it may be seen that this agrees fairly well with that predicted. No details of the performance of the Craigneuk moulds is available.

It is apparent that if foundries could modify the manufacturing processes to give similar properties as , say, Distington material then some improvement in performance may be possible. For example, the Dowlais and Landore carbon

contents are significantly lower than those at Distington (Table 70). If these were raised, improved cracking resistance may be obtained.

CHAPTER SIX

IMPLICATIONS TO INGOT MOULD PERFORMANCE

6.1 INGOT MOULD ANALYSIS

The two critical failure modes in flake graphite ingot moulds are cracking and crazing. It has been shown that if the onset of cracking is delayed or eliminated, so that the mould is rejected due to crazing, then mould performance is improved. It is common experience that this may be achieved by increasing the fraction of the graphitising elements, such as carbon and silicon, and reducing the pearlite stabilising elements, such as manganese. The effect of this will be to produce an iron with an increased volume fraction of ferrite and coarser graphite, so that, although the iron becomes weaker, its ductility is increased. However, these changes, if applied to a mould failing by crazing, would be likely to cause a decrease in mould life, since the crazing process is accelerated by these effects.

The present work provides an evaluation of the effect of individual elements on mould performance as follows.

6.1.1 Carbon

It was shown, in a consideration of the performance of large (25t) slab moulds, that carbon exerted a significant influence. The addition of broken mould scrap caused a reduction in carbon content from 3.96% to 3.91% and this was associated with a drop in performance. Both changes were statistically significant and regression analysis generated the following equation ;-

$$LIFE = -196.7 + 65.60 \times (\%C).$$

It is apparent, therefore, that in large slab moulds, increased carbon levels improve performance and this is attributed to increasing the volume fraction of graphite, making the iron softer and more ductile and, hence, improving

the cracking resistance.

Carbon levels in Dowlais and Landore moulds are significantly lower than Distington moulds. The properties of the iron produced at these foundries is inferior in terms of cracking resistance and the mould performances are worse.

Of the smaller square moulds failing by crazing, no evidence of a correlation between performance and carbon content has been found. It may be deduced that carbon will exert a similar effect on cracking resistance and so it should be maintained at the current 3.7-3.9% level.

6.1.2 Silicon

Silicon was found to behave in a similar manner to carbon, in large slab moulds, that is, increased levels improving performance by reducing cracking. In addition, it has been argued that high silicon levels in forging moulds promote surface plucking by weakening the iron.

Silicon, therefore, in the range 1-2%, is considered to improve cracking resistance of iron by increasing the the volume fraction of ferrite and by its graphitising action.

High silicon levels, say at around 5%, promote the formation of type D undercooled graphite, and bend test data have indicated that this would have a highly detrimental effect on cracking resistance. The effect of intermediate levels (2-5%) on flake graphite remain unsure from the present work, but it may be anticipated that the beneficial effects on cracking resistance from ferritisation and graphitisation will be increasingly offset by the embrittling effect of solid solution strengthening.

6.1.3. Manganese.

Manganese is a pearlite stabilising element, and as such, its effect will be opposite to carbon and silicon; that is, reduced manganese levels are required to improve iron ductility and reduce cracking.

6.1.4. Sulphur.

It is known that low sulphur levels are detrimental to iron inoculation, in the case of engineering grey irons. Several River Don forging moulds have been produced with sulphur levels below 0.01%. To date, no detrimental effects of these low sulphur levels have been observed, although data are limited. Cooling rates in ingot moulds are too slow to allow the formation of vermicular graphite at low sulphur levels. There is no evidence to suggest that sulphur has a significant effect on mould performance in levels up to 0.1%.

6.1.5. Phosphorus.

Phosphorus in ingot mould irons forms a low melting (960°) ternary carbide/phosphide eutectic. At normal levels of around 0.06% P, the eutectic appears in the form of isolated pools. At levels above 0.1%, however, the eutectic starts to form as a cell boundary network and this morphology is considerably enhanced by fast cooling rates and high residual element concentrations. It has been shown that when the eutectic is in an isolated form, it has little effect on mechanical properties, but when segregated to the cell boundaries, causes a marked embrittlement. Such embrittlement has been shown to cause a drop in performance of small, square moulds, by promoting premature cracking.

It has been noted that high phosphorus levels (say, 0.2%) in themselves, are not necessarily detrimental to performance, but that, when accelerated cooling occurs, levels below 0.12% are necessary, to prevent cracking. The importance of stripping practice will be discussed, below.

6.1.6. Titanium.

In irons of ingot mould composition, it has been shown that up to 0.2% titanium acts as a graphitiser, increasing the graphite coarsness and it is also mildly ferritising. Because of these effects, the strength of the iron is reduced but the iron is more ductile. The effect of these factors on ingot moulds is to reduce the incidence of cracking and promote crazing and, also, by reducing the strength, to increase failures by torn seats and broken lugs. The overall effect on performance of large slab moulds has been found to be, initially, an increase in life with up to 0.04-0.06% titanium, by improving the cracking resistance. This is followed by a reduction in life, with up to 0.1% titanium, by increasing failures at low life, through torn seats, etc. In large forging moulds, high titanium levels (0.06%) have been associated with excessive weakness resulting in surface plucking.

At levels of 0.2% titanium and above the formation of type D undercooled graphite has been observed. This has been found to be associated with marked embrittlement and would be expected to give poor mould performance, due to premature cracking.

The redox conditions employed during melting have been found to play a significant role. At a given titanium level, it has been established that oxidising conditions (for

example by teatment wih carbon dioxide) result in finer graphite stuctures, whereas reducing conditions (for example by treatment with hydrogen) produce coarser graphite structures.

At low titanium levels, a ductility trough occurs, typically in the range 0.03 to 0.06% titanium. The precise location of the ductility trough varies with melt redox condition being at higher titanium levels in oxidising conditions. There is some evidence to suggest that the performance of moulds produced within this ductility trough are inferior to other moulds, but more investigation is required to confirm this effect.

It has been observed that if oxidised scrap is used in electric melting, then the optimum titanium level may be different to that in cupola melted iron.

Historically, titanium has been added to iron to improve mould performance, when it has been considered that performance is poor due to nitrogen compaction. This is discussed more fully in the next section on the effects of nitrogen.

6.1.7 Nitrogen

Nitrogen levels in experimental test block material of ingot mould composition have been increased up to 0.018%, but no evidence for the formation of compacted graphite has been found. Up to this level, little or no effect on mechanical properties was observed. It was noted, however, that above 0.012% nitrogen pinholes were formed, and this was considered to be detrimental to performance by forming a region of low thermal conductivity behind the hot face of the mould.

Considerable evidence has been presented, from actual ingot moulds, to support this view that nitrogen does not cause compaction and is not detrimental to mould performance. In addition, it has also been argued that the presence of compacted graphite is not detrimental to mould performance.

It has been shown that when hot blast cupolas melting 100% steel scrap were introduced in the 1960's and performances were poor, the action of adding titanium to fix nitrogen, and, hence, prevent the formation of compacted graphite, was effective because of the efficacy of titanium as a graphitiser. It is likely the residual levels would also have been increased by replacing pig iron with steel scrap, and this would have caused a reduction in performance (see next section). The beneficial effect of titanium was, therefore, in counteracting the effects of increased residual levels.

6.1.8 Residual Elements

Experimental castings have shown that increased residual element levels cause a marked increase in brittleness. This occurs, initially, by increasing the pearlite content of the matrix and, subsequently, by forming carbides. At high residual levels typical of the maximum recommended by the Code of Practice, the carbides become increasingly segregated to the cell boundaries, offering easy crack propagation paths.

In large slab moulds, cracking resistance is of greatest importance and it may be expected that residual contents will play a major role in these moulds. No direct evidence of this has been found but it is noteworthy that Distington foundry operated with the lowest residual levels compared

with the other foundries and their mould performance is generally considered to be superior.

The presence of excessively high residual levels, particularly chromium, has been cited as the cause of premature failures. High chromium levels are evident in the form of cell boundary carbide, which promotes premature cracking.

Increased residual levels have been shown to promote the possibly deleterious effects of high phosphorus levels on performance.

It has been shown that increased residual levels promote the formation of compacted graphite, even at relatively low nitrogen levels, although the precise mechanism for the formation of compacted graphite remains unclear.

6.2 FOUNDRY PRODUCTION PARAMETERS

6.2.1. Running Systems And Metal Cleanness.

The experience of River Don forging moulds has demonstrated that the accidental entrapment of non-metallic inclusions could cause a decrease in mould performance, in this instance, by promoting surface plucking. This entrapment may be reduced by the following;-

i) Ensure sound brickwork in the ladle.

ii) Efficient ladle deslagging.

iii) Design of the running system, to give smooth entry, thus avioding mould wall erosion.

iv) Use of ingate design to prevent dross entrapment.

v) Use of whirlgates.

6.2.2. Casting Temperature.

Reduced casting temperatures, down to 1200°C, have been shown to produce a more ductile iron, by reducing the eutectic cell size. However, casting temperatures as low as this could give increased horizontal cracking in slab moulds, due to the incidence of cold shuts, and decreased performences in small, square moulds failing by crazing. The casting temperature should, therefore, be maintained at 1260-1280°C.

6.2.3. Cooling Rates And Stripping Practice.

Reduced cooling rates in low residual melts have been shown to improve cracking resistance. However, the cooling rate range so far examined is higher than normally found in ingot mould castings. It is considered, though, that in moulds failing by cracking, such as large slab moulds, the cooling rate should be as slow as possible. The position of the higher residual small, square mould failing by crazing remains unclear, but it is suggested that these moulds should be, also, cooled slowly. Fast cooling following premature sand removal at temperatures of 1000°C, has been shown to be highly detrimental to performance by promoting brittleness and, hence, mould cracking. It has also been shown that exposure of any part of the casting may cause local embrittlement, also leading to premature failures. Moulds should be stripped below 700°C, only.

By a detailed examination of experimental material and production ingot moulds, the following conclusions with regard to the effect of production variables on mould performance may be made

Mould Analysis

1) Carbon levels should be maintained at above 3.7%.

2) High silicon levels in forging moulds, up to 2.02%, promote surface plucking.

3) Reduced carbon equivalent (4.46% down to 4.40%) promotes premature cracking in slab moulds.

4) High phosphorus levels(0.12-0.3%) may be detrimental to mould performance when associated with fast cooling.

5) Titanium acts as a graphitiser and ferritiser in ingot mould iron up to 0.2%. There is an optimum titanium content, below which the incidence of cracking increases and above which, crazing and torn seats result. In a large slab mould, this has been shown to be around 0.05%.

6) Titanium causes a ductility trough at 0.03-0.05% Ti.

7) The effects of titanium are markedly influenced by treatment with carbon dioxide and hydrogen. Melts produced from used ingot mould scrap behave as if they were treated with carbon dioxide.

8) Nitrogen levels above 0.014% promote pinholing.

9) There is little correlation between nitrogen levels up to 0.018% and graphite compaction.

10) Compacted graphite is not, necessarily, detrimental to ingot mould performance.

11) High residual levels (0.03% Mo, 0.14% Cr, 0.025% Sn, 0.03% V, 0.08% Ni, 0.3% Cu and 0.008% As) cause a reduction in cracking resistance.

12) High residual levels promote the segregation of carbide/phosphide eutectic to eutectic cell boundaries.

13) High residual levels promote the formation of compacted graphite.

Mould Production.

14) Reduced casting temperatures, from 1300°C down to 1200°C, reduce eutectic cell size and increase cracking resistance.

15) Casting temperatures below 1250°C are responsible for lower life in slab moulds, due to the increased incidence of cold shuts.

16) Increased casting temperatures of up to 1340°C improve the performance of small, square moulds failing, principally, by crazing.

17) Additions of 0.2% FeSi in heavy sections reduce eutectic cell size and increase cracking resistance.

18) With low residual levels, slower cooling rates reduce brittleness, by increasing ferrite and graphite coarseness.

19) With high residual levels, slower cooling rates increase brittleness, by allowing greater segregation.

20) Stripping of the sand jacket at temperatures above 900°C decreases mould performance, by promoting the segregation of carbides to eutectic cell boundaries.

Mould Microstructure.

21) Increasing ferrite content increases ductility.

22) Increasing graphite coarseness increases ductility.

23) The formation of undercooled graphite is markedly detrimental.

24) Random distributions of carbide phases are not,

necessarily, detrimental.

25) Segregation of carbides and carbide/phosphide eutectic to eutectic cell boundaries causes a marked increase in brittleness.

Mould Mechanical Properties.

26) The cracking resistance of ingot mould iron may be adequately characterised by a three-point bend test.

27) Properties may be monitored on a routine basis, using lug samples.

28) Different mould foundries produce inherently different mould properties.

29) Properties may vary within an ingot mould; for example, due to localised variations in cooling rate.

8.1 CARBON AND SILICON

The present work demonstrated that a drop in carbon content and carbon equivalent as small as 0.05% apparently had a large effect on slab mould life. Statistical studies of a larger mould population are required to determine whether this effect is real or whether it is due to a number of moulds with very low carbon equivalent values failing prematurely.

It has been suggested that increased silicon levels above 2.0% would give increased brittleness due to solid solution strengthening. This aspect requires to be quantified.

8.2 TITANIUM

Both manganese and titanium form sulphides, thus the effects of titanium may vary with manganese and sulphur contents. The effects of the interaction of these elements on cracking resistance require to be studied in more detail.

The effects of the observed ductility trough on ingot mould performance require to be studied. This may be carried out by a statistical examination, but a larger mould population than available during the course of this work is required. The mechanism for the formation of the ductility trough should be further explored.

Titanium is added to quasi-flake irons to control graphite morphology. Since titanium has been found to have such significant effects in flake iron, this should be pursued in quasi-flake irons.

8.3 RESIDUAL ELEMENTS

The detrimental effects of high residual element levels on cracking resistance have been established in experimental

material. Confirmation of the expected decrease in the performance of large slab moulds is now required.

8.4 COMPACTED GRAPHITE

The contribution of high residual element levels to the formation of compacted graphite has been demonstrated. This work should be expanded to compare the effect of high residuals at both low and high nitrogen content. The effect of section size, and hence cooling rate, should also be included. The possible effects of ferrostatic pressure may also be assessed by comparing the structure at the top and bottom of tall castings.

8.5 COOLING RATE

The present method of defining cooling rates based on sand thickness is, clearly, inadequate. Some attempt should be made to quantify the effects of cooling rate on cracking resistance in terms of, say, a maximum cooling rate at a given temperature. This could then be used to specify the optimum sand thickness for a given mould/tackle system.

8.6 LUG SAMPLES

Ingot mould performance from a particular foundry can drop over a period of time but the reasons for this are usually extremely difficult to determine. The provision of large lug samples on every one in ten moulds, say, would allow some check on possible changes in mechanical properties with time. Before this stage is reached, however, it is considered that further studies of its suitability be carried out.

8.7 HOLDING TIMES

Iron may be held for long periods of time, for example over a fortnight's holiday, in an electric furnace. The

effects of this on ingot mould performance should be established by means of a statistical examination.

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Table 1 Typical Analyses of Raw Materials Used for Experimental Melts

Source	Warner Iron	Bremanger Iron	Renishaw Mould Scrap
C	3.8	3.8	3.75
Si	0.5-1.0	0.5-1.0	1.2
S	0.08	0.015	0.05
P	0.06	0.02	0.05
Mn	0.50	0.20	0.7
Mo	0.02	0.002	0.01
Cr	0.08	0.03	0.02
Sn	0.02	0.001	0.001
V	0.02	0.001	0.01
Ti	0.02	0.001	0.02
Ni	0.06	0.001	0.006
Cu	0.10	0.03	0.05

Table 2 Measured And Derived Parameters From The Three-Point Bend Test (see figure 4)

Nomenclature	Derivation	Significance
M_L	Maximum load measured directly from plot	Related to tensile strength
A	Area under curve up to maximum load	Proportional to energy to initiate fracture
B	Area under curve after maximum load	Proportional to energy to propagate fracture
A/B	Ratio of A/B	High value indicates brittle material
a	Deflection up to maximum load at half peak height	
b	Deflection after maximum load at half peak height	
a/b	Ratio of a/b	High value indicates brittle material
$A_{\frac{1}{2}}$	Area under curve up to maximum load at half peak height	
$B_{\frac{1}{2}}$	Area under curve after maximum load at half peak height	
$A\frac{1}{2}/B\frac{1}{2}$	Ratio of $A\frac{1}{2}/B\frac{1}{2}$	High value indicates brittle material
A_n	A/M_L	Normalised initiation energy
B_n	B/M_L	Normalised propagation energy
A+B	Total area under curve	High values indicate tougher material, i.e. total energy to produce failure

Table 3 Effect of Graphite Size On Bend Test Parameters

Sample No.	Max. Load, kN	a/b	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	A	B	A/B	A_n	B_n	A+B
1	3.5	35	200	2.2	0.01	200	0.63	0.003	2.2
2	2.1	20	20	1.3	0.2	5.8	0.62	0.10	1.5
3	1.5	9.2	8.1	1.2	0.5	2.8	0.80	0.33	1.7
4	1.4	2.1	2.3	1.1	0.8	1.4	0.79	0.57	1.9
5	1.4	2.3	2.5	1.0	1.1	0.9	0.71	0.79	2.1

Table 4 Effect of Percentage Pearlite On Bend Test Parameters

Pearlite, %	Max. Load, kN	a/b	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	A	B	A/B	A_n	B_n	A+B
100	1.6	2.9	3.9	1.0	0.7	1.5	0.63	0.44	1.7
40	1.3	2.7	3.9	1.0	0.8	1.4	0.77	0.62	1.8
0	1.3	1.7	2.4	1.0	1.0	1.0	0.77	0.77	2.0

Table 5 Effect Of Carbides In Flake Iron On Bend Test Parameters

Condition	Max. Load, kN	a/b	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	A	B	A/B	A_n	B_n	A+B
Carbide free	1.6	3.5	4.3	1.5	0.6	2.5	0.94	0.38	2.1
Carbidic	2.5	17.5	15.7	2.0	0.3	6.7	0.80	0.12	2.3

Table 6 Effect Of Graphite Morphology On Bend Test Parameters

Condition	Max. Load, kN	a/b	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	A	B	A/B	A_n	B_n	A+B
Ferritic Hematite	1.3	1.7	2.4	1.0	1.0	1.0	0.77	0.77	2
Compacted	2.6	2.6	3.5	3.5	2.0	1.7	1.37	0.79	3.6
Ferritic Q/F	3.5	3.5	3.4	7.4	2.9	2.6	2.14	0.83	10.4

Table 7 Effect Of Carbides In Quasi-Flake iron On Bend Test Parameters

Condition	Max. Load, kN	a/b	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	A	B	A/B	A_n	B_n	A+B
Carbide free	3.5	3.5	3.4	7.5	2.9	2.5	2.1	0.8	10.4
Carbidic	4.0	10.9	11.5	9.7	1.6	6.3	2.4	0.4	11.4

Table 8 Chemical Analysis of Test Blocks To Determine The Effects Of Phosphorus

Cast No.	Strip Temp.	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Cu
RG202/A	>1000	3.75	1.80	0.012	0.058	0.46	0.001	0.001	0.001	0.010	0.001	0.029	0.03
RG202/B	< 700												
RG203/A	>1000	3.95	1.74	0.011	0.317	0.73	0.001	0.001	0.001	0.010	0.001	0.024	0.03
RG203/B	< 700												
RG742/A	>1000	3.61	1.30	0.031	0.058	0.89	0.01	0.01	0.009	0.010	0.013	0.05	0.06
RG742/B	>1000	3.55	1.18	0.027	0.325	0.79	0.01	0.01	0.008	0.015	0.017	0.05	0.07
RG743/A	< 700	3.60	1.31	0.025	0.059	0.91	0.01	0.01	0.010	0.010	0.013	0.05	0.07
RG743/B	< 700	3.63	1.23	0.024	0.310	0.94	0.01	0.01	0.010	0.015	0.019	0.05	0.07

Cast No.	As	Co	N	Al
RG202/A	-	-	-	0.012
RG202/B				
RG203/A	-	-	-	0.018
RG203/B				
RG742/A	0.010	0.016	0.007	0.005
RG742/B	0.010	0.017	0.007	0.005
RG743/A	0.011	0.024	0.007	0.005
RG743/B	0.011	0.013	0.007	0.005

Table 9 Mechanical Properties Of Test Blocks With Varying Phosphorus Levels

Cast No.	Residual Level	Ferrite %	Strip Temp. °C	P	UTS N/mm^2	Elong. %	Max Load kN	a/b	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	A	B	A/B
RG202/B	Low	85	< 700	Low	39	0	1.0	1.1	1.6	0.8	1.0	0.8
RG202/A	Low	35	>1000	Low	59	1.4	1.2	2.3	2.6	1.1	1.0	1.1
RG203/B	Low	35	< 700	High	64	1.3	1.2	1.8	2.4	1.0	1.1	0.9
RG203/A	Low	5	>1000	High	85	0	1.6	2.9	3.9	1.0	0.7	1.5
RG743/A	High	5	< 700	Low	118	1.6	1.6	3.5	4.3	1.5	0.6	2.5
RG742/A	High	5	>1000	Low	144	0.6	1.9	7.1	6.9	1.8	0.8	2.3
RG743/B	High	5	< 700	High	177	1.3	2.1	14.3	8.9	1.8	0.4	4.5
RG742/B	High	5	>1000	High	213	OPM	2.5	17.5	15.7	2.0	0.3	6.7

OPM - Outside punch marks

Table 10 Chemical Analysis of Light Section Melt To Determine Effects Of Titanium

Cast No.	C	Si	S	P	Mn	Cr	Mo	Ni	Al	Cu	Sn	V	Ti
RG480	3.75	2.02	0.038	0.032	1.3	0.17	0.010	0.020	0.008	0.03	0.003	0.054	0.022

Table 11 Chemical Analyses Of The Light Section Test Blocks

Ti Addition Rate	Gas Treatment	% Ti	% N	
			Sol.	Insol.
Nil	H_2	0.02	0.0016	0.0028
Medium	H_2	0.18	0.0016	0.0030
High	H_2	0.35	0.0016	0.0010
Nil	CO_2	0.03	0.0016	0.0042
Medium	CO_2	0.24	0.0014	0.0024
High	CO_2	0.39	0.0022	0.0012
Nil	None	0.03	0.0020	0.0046
Medium	None	0.20	0.0020	0.0048
High	None	0.41	0.0018	0.0024

Table 12 Chemical Analyses of Heavy Section Test Blocks To Determine The Effects Of Titanium

Cast No.	C	Si	S	P	Mn	Cr	Mo	Ni	Al	Cu	Sn	V	Ti	N
RG300	3.78	1.67	0.034	0.030	1.08	0.10	0.002	0.032	0.006	0.03	0.001	0.006	0.023	0.007
RG305	3.71	1.48	0.039	0.030	1.14	0.10	0.007	0.037	0.002	0.03	0.001	0.003	0.027	0.005
RG301	3.79	1.57	0.033	0.028	1.02	0.06	0.001	0.012	0.004	0.03	0.001	<0.001	0.036	0.006
RG340	3.73	1.73	0.035	0.030	1.07	0.10	0.004	0.035	0.006	0.03	0.001	0.011	0.079	0.005
RG341	3.76	1.60	0.035	0.029	1.05	0.10	0.004	0.033	0.007	0.03	0.001	0.007	0.196	0.004
RG344	3.75	1.63	0.035	0.029	1.04	0.08	0.050	0.032	0.003	0.03	0.001	0.014	0.113	0.005
RG718	3.76	1.43	0.025	0.024	0.90	0.02	0.01	0.04	ND	0.05	0.011	0.01	0.041	0.003
RG724	3.62	1.49	0.024	0.024	0.89	0.03	0.005	0.05	ND	0.07	0.008	<0.01	0.046	ND
RG769	3.70	1.54	0.015	0.022	0.84	0.02	0.005	0.04	ND	0.06	0.008	0.01	0.012	0.006
RG754	3.63	1.15	0.022	0.022	0.83	0.01	0.005	0.04	ND	0.06	ND	0.017	0.27	0.004
RG725	3.64	1.40	0.022	0.022	0.90	0.03	0.005	0.05	ND	0.06	0.012	<0.01	0.027	ND
RG719	3.69	1.36	0.024	0.023	0.89	0.02	0.01	0.04	ND	0.05	0.010	<0.01	0.056	0.004
RG770	3.62	1.61	0.027	0.020	0.83	0.02	0.005	0.05	ND	0.06	0.012	<0.01	0.10	0.005
RG775	3.67	1.12	0.023	0.022	0.83	0.02	0.008	0.04	ND	0.06	ND	0.011	0.24	0.005

Table 13 Bend Test Properties Of Light Section Test Blocks With Various Titanium And Gas Treatments

Gas Treatment	Ti	Max. Load, kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A	B	A/B
H_2	0.02	1.5	9.2	8.1	1.2	0.5	2.8
	0.18	1.7	2.3	2.9	1.1	0.8	1.3
	0.35	2.1	~20	~20	1.3	0.2	5.8
None	0.03	1.6	6.1	7.5	1.0	0.5	2.5
	0.20	2.3	~20	~15	1.5	0.5	3.0
	0.41	3.0	~30	~50	1.5	0.01	~60
CO_2	0.03	1.6	5.3	4.6	1.0	0.5	2.0
	0.24	3.5	~35	~100	2.2	0.01	~100
	0.39	3.6	~40	~100	2.1	<0.01	>100

Table 14 Mechanical Properties Of Heavy Section Test Blocks With Various Titanium And Gas Treatments

Cast No.	Gassing Treatment	%Ti	Tensile Test		Bend Test					
			UTS, N/mm^2	Elong.%	Max. Load, kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A	B	A/B
RG718	H_2	0.041	87	1.4	1.5	2.4	2.9	1.2	0.7	1.9
RG724	"	0.046	69	1.1	1.3	1.8	2.1	0.9	0.8	1.1
RG769	"	0.12	66		1.3	2.2	2.8	1.3	1.0	1.3
RG754	"	0.24	98		1.8	1.8	2.2	1.6	1.5	1.0
RG300	None	0.023	79.5	1.20	1.4	2.2	2.3	1.0	0.8	1.3
RG305	"	0.027	90.5	1.20	1.3	1.9	2.7	1.0	1.0	1.0
RG301	"	0.036	70.0	1.40	1.6	4.7	6.0	1.4	0.7	2.0
RG340	"	0.079	65.0	1.00	1.4	2.1	2.6	1.2	1.1	1.1
RG344	"	0.113	74.0	1.15	1.4	2.3	2.5	1.0	1.1	0.9
RG341	"	0.196	82.5	1.25	1.9	3.7	4.2	1.8	1.8	1.0
RG725	CO_2	0.027	65	1.0	1.3	2.5	2.9	1.0	0.8	1.3
RG719	"	0.056	95	1.1	1.7	3.0	3.3	1.7	0.9	2.0
RG770	"	0.10	77		1.5	1.8	2.1	1.4	1.4	1.0

Table 15 Chemical Analyses Of Test Blocks Melted Using Clean And Dirty Scrap

Cast No.	Scrap Condition	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Al
RG801	Clean	3.7	1.00	0.052	0.051	0.67	<.01	0.02	0.005	<.01	0.036	0.008	0.005
RG802	Clean	3.8	0.99	0.049	0.048	0.68	<.01	0.02	0.004	<.01	0.049	0.005	0.005
RG803	Dirty	3.5	0.82	0.042	0.061	0.50	<.01	0.01	0.003	<.01	0.037	0.01	0.005
RG804	Dirty	3.5	0.98	0.038	0.048	0.52	<.01	0.01	0.002	<.01	0.061	0.006	0.005

Table 16 Mechanical Properties Of Test Blocks Melted Using Clean And Dirty Scrap

Cast No.	Scrap Condition	%Ti	UTS N/mm^2	Max. Load kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A	B	A/B
RG801	Clean	0.036	92.3 and 92.9	1.7	3.5	3.3	1.3	0.8	1.6
RG802	Clean	0.049	89 and 92.9	1.5	3.2	4.0	1.3	0.7	1.8
RG803	Dirty	0.037	116 and 122	1.9	3.6	4.7	1.4	0.6	2.4
RG804	Dirty	0.06	117 and 121	1.8	3.65	3.95	1.3	0.7	1.9

Table 17 Treatments Used During Experiments To Determine The Effects Of Nitrogen

Cast No.	Treatment	Stripping Practice
RG666A	0.5% $CaCN_2$ + 0.5% Soda Ash	<700°C
RG751A	1.0% $CaCN_2$ + 1.0% Soda Ash	<700°C
RG752A	2.0% $CaCN_2$ + 2.0% Soda Ash	<700°C
RG666B	0.5% Calsiloy + 0.5% Soda Ash	<700°C
RG751B	1.0% Calsiloy + 1.0% Soda Ash	<700°C
RG752B	2.0% Calsiloy + 2.0% Soda Ash	<700°C
RG667A	High Nitrogen Ferromanganese	>1000°C
RG667B	High Nitrogen Ferromanganese	<700°C
RG668A	N_2 Bubbled At 1600°C, 10 mins	>1000°C
RG668B	N_2 Bubbled At 1600°C, 10 mins	<700°C

Table 18 Test Block Analyses To Determine The Effects Of Nitrogen

Cast No.	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Al	Ca	N
RG666A	3.63	1.66	0.024	0.021	0.91	0.003	0.02	0.010	0.005	0.011	0.05	0.009	<0.001	0.0054
RG751A	3.79	1.51	0.037	0.028	0.82	0.007	0.05	0.011	0.010	0.009	0.04	<0.01	<0.001	0.014
RG752A	3.80	1.46	0.017	0.027	0.82	0.005	0.03	0.011	0.012	0.009	0.04	<0.01	<0.001	0.018
RG666B	3.63	1.59	0.018	0.022	0.92	0.003	0.02	0.012	0.005	0.011	0.05	0.008	0.0012	0.0072
RG751B	3.65	1.89	0.019	0.033	0.83	0.007	0.04	0.008	0.010	0.010	0.04	<0.01	0.0020	0.003
RG752B	3.57	2.26	0.011	0.030	0.81	0.008	0.03	0.005	0.013	0.009	0.04	<0.01	0.0035	0.003
RG667A	3.57	1.81	0.020	0.021	0.92	0.003	0.02	0.010	0.005	0.012	0.05	0.006	0.0003	0.0122
RG667B	3.70	1.72	0.024	0.020	0.86	0.003	0.02	0.010	0.008	0.009	0.05	0.007	0.0003	0.0112
RG668A	3.65	1.63	0.026	0.022	0.87	0.003	0.02	0.012	0.008	0.010	0.05	0.012	0.0004	0.0048
RG668B	3.69	1.51	0.026	0.021	0.83	0.003	0.02	0.011	0.009	0.010	0.05	0.028	0.0004	0.0050

Table 19 Changes In Nitrogen And Calcium Levels On Treatment

Cast No.	Addition	N		Ca	
		Bath	Final	Bath	Final
RG666A	0.5% $CaCN_2$	0.0062	0.0054	0.0001	<0.001
RG751A	1% $CaCN_2$	0.007	0.014	0.0006	<0.001
RG752A	2% $CaCN_2$	0.005	0.018	0.00013	<0.001
RG666B	0.5% Calsiloy	0.0062	0.0072	0.0001	0.0012
RG751B	1% Calsiloy	0.007	0.003	0.0006	0.0020
RG752B	2% Calsiloy	0.005	0.003	0.00013	0.0035
RG667A	NFeMn	-	0.0122	-	0.0003
RG667B	NFeMn	-	0.0112	-	0.0003
RG668A	N_2 Gas	0.0066	0.0048	0.0004	0.0004
RG668B	N_2 Gas	0.0066	0.0050	0.0004	0.0004

Table 20 Mechanical Properties Of Nitrogen Treated Test Blocks

Cast No.	N	Ca	UTS N/mm^2	Max. Load kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A/B	A	B
RG666A	0.0054	0.001	120	1.86	6.7	4.5	2.9	1.4	0.5
RG751A	0.014	0.001	127.5	2.16	6.0	5.5	2.8	1.5	0.6
RG752A	0.018	0.001	113	2.00	5.8	5.6	2.1	1.1	0.5
RG666B	0.0072	0.0012	109	1.99	3.7	4.3	1.9	1.3	0.7
RG751B	0.003	0.0020	114	1.90	3.9	4.1	1.8	1.4	0.8
RG752B	0.003	0.0035	73	1.21	1.8	2.2	1.0	0.6	0.6
RG667A	0.0122	0.0003	114	1.86	3.9	4.5	1.7	1.1	0.7
RG667B	0.0112	0.0003	107	1.78	3.6	4.4	2.0	1.5	0.7
RG668A	0.0048	0.0004	119	1.88	5.4	5.7	2.2	1.2	0.6
RG668B	0.0050	0.0004	98.4	1.68	2.7	4.0	1.7	1.2	0.8

Table 21 Analyses Of The Test Blocks With Varying Residual Levels

Cast No.	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Cu	As	N
RG720 A & B	3.70	1.35	0.024	0.026	0.88	0.004	0.03	0.010	-	0.011	0.04	0.06	-	-
RG852	3.72	1.36	0.018	0.018	0.85	0.005	0.01	0.012	0.005	0.007	0.05	0.06	0.006	0.00
RG853	3.76	1.29	0.019	0.017	0.80	0.005	0.01	0.010	0.006	0.007	0.05	0.07	0.006	0.00
RG854 & B	3.74	1.21	0.083	0.061	0.76	0.020	0.09	0.026	0.015	0.018	0.06	0.11	0.007	0.00
RG865	3.80	1.16	0.081	0.056	0.80	0.021	0.10	0.023	0.019	0.017	0.06	0.10	0.009	0.01
RG867 A & B	3.50	1.38	0.079	0.059	0.84	0.032	0.14	0.026	0.03	0.02	0.07	0.29	0.008	0.01
RG868	3.70	1.49	0.041	0.073	0.79	0.033	0.12	0.025	0.03	0.02	0.08	0.30	0.007	0.01

Table 22 Mechanical Properties Of The Test Blocks With Varying Residual Levels

Cast No.	Test Block Size mm	UTS N/mm^2	Max. Load kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A/B	A	B	Residual Level
RG720A	254	123	1.83	3.5	4.3	1.7	1.2	0.8	Low
RG720B	254 + sleeve	135	1.9	3.0	3.8	1.7	1.4	0.8	
RG853	305 + sleeve	108.5	1.82	2.4	3.0	1.6	1.2	0.7	
RG852	355 + sleeve	102	1.15	2.7	3.0	1.9	0.9	0.4	
RG854A	254	113	2.14	4.5	4.4	1.9	0.9	0.4	Medium
RG854B	254 + sleeve	119.5	1.91	5.0	6.0	2.5	1.1	0.6	
RG865	305 + sleeve	129.5	2.40	5.3	5.3	2.1	1.7	0.9	
RG867A	254	118.5	2.05	4.7	-	-	-	-	High
RG867B	254 + sleeve	116.5	2.20	5.0	5.5	2.0	1.3	0.7	
RG868	305 + sleeve	123	2.6	5.7	6.2	2.1	1.8	0.9	

Table 23 Chemical Analyses Of Test Blocks To Determine The Effect Of Casting Temperature

Cast No.	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Al	Cu
RG573/A & B	3.63	1.62	0.037	0.053	0.75	0.010	0.029	0.005	0.007	0.005	0.027	0.004	0.035
RG574/A	3.69	1.57	0.036	0.053	0.79	0.009	0.028	0.005	0.007	0.003	0.027	0.002	0.033
RG574/B	3.67	1.65	0.035	0.027	0.76	0.009	0.017	0.004	0.001	0.002	0.027	0.003	0.031

Table 24 Macrostructure And Mechanical Properties Of Test Blocks With Varying Casting Temperature

Cast No.	Casting Temperature °C	Pearlite Interlamellar Spacing, Å	Eutectic Cell Size cm	Tensile Strength N/mm	Bend Test Properties					
					Max. Load kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A	B	A/B
RG574/A	1300	3800	0.65	66	1.3	3.3	4.1	0.9	0.6	1.6
RG573/A	1250	5300	0.55	68	1.1	2.0	2.5	0.8	0.6	1.2
RG573/B	1200	4200	0.50	63	1.1	2.0	2.4	0.8	0.7	1.2
RG574/B	1250 + FeSi	5200	0.45	71	1.1	1.8	2.0	0.7	0.7	1.0

Table 25 Effect Of Superheat-Analyses Of The Test Blocks

Cast No.	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Al	Cu
RG869	3.70	1.22	0.056	0.058	0.85	0.005	0.01	0.013	0.01	0.01	0.05	0.005	0.06
RG903	3.61	1.74	0.052	0.054	0.84	0.010	0.04	0.014	0.010	0.009	0.06	0.045	0.06

Table 26 Effect Of Superheat On Mechanical Properties

Cast No.	Superheat Temperature °C	Tensile Strength N/mm^2	Bend Test Properties					
			Max. Load kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A	B	A/B
RG869A	1300	76	2.0	3.5	3.9	1.7	0.8	2.0
RG869B	1400	111	2.0	3.1	3.9	1.1	0.7	1.7
RG903A	1500	125	1.9	4.0	5.5	1.0	0.5	2.0
RG903B	1600	122	2.1	2.8	2.9	1.3	0.8	1.9

Table 27 Effect Of Cupola Burden On 48-Type Mould Performance

Burden	No.	Mean	Max	Min	SD	Date Of Production
70/20/10	155	67.88	120	11	21.76	Mid 1974-End 1975
53/17/20/10	46	54.94	100	1	23.85	Mid 1975-June 1976
63/17/20	67	63.34	106	14	21.35	June 1976-Nov. 1976
63/17/20+FeTi	63	65.03	101	2	23.60	Nov.-

Table 28 t-Test For Significance Of Performance Of 48-Type Moulds Between Burdens (data as at 20/1/78)

	Degrees Of Freedom	Combined σ	t	Level Of Significance
70/20/10 v 53/17/30	252	21.66	6.71	99.9%
53/17/30 v 63/17/20	164	21.45	4.17	99.9%
63/17/20 v 63/17/20+Ti	116	22.23	0.20	NS
70/20/10 v 63/17/20	220	21.64	1.44	NS
70/20/10 v 63/17/20+Ti	204	22.16	1.50	NS

Table 29 Amount Of Repairing On 48-Type Moulds

Burden	No. Of Moulds Failed	No. Of Repairs To Failed Moulds	% Repaired
70/20/10	155	79	51
53/17/20/10	100	24	24
62/17/20	67	14	21
63/17/20 + Ti	51	10	20

Table 30 Mode Of Failure Of 48-Type Moulds

Cupola Burden	Reason For Failure								
	Crazing %	Vertcal Cracks %	Horizontal Cracks %	Torn Seats (Teeming) %	Broken Lug %	Mechanical %	Broken Balling %	Cold Shut %	Torn Seat (Foundry %
70/20/10	11.8	66.9	10.1	2.4	3.0	0.6	3.0	1.8	
53/17/20/10	1.0	75.5	1.0		1.0		2.0	17.3	1.0
63/17/20		82.1	3.0		1.5		11.9	1.5	
63/17/20 + Ti		86.3			2.0		3.9	2.0	5.9

Table 31 Position Of Vertical Cracks In 48-Type Moulds

Cupola Burden	Crack Position		
	Broad Side, %	Narrow Side, %	Corner, %
70/20/10	57.7	35.8	7.3
53/17/20/10	44.6	48.9	6.8
63/17/20	69.1	25.5	5.5
63/17/20 + Ti	76.6	24.4	0

Table 32 Production Details Of Fullwood 48-Type Moulds

Variable	70/20/10					53/17/20/10				
	No.	Mean	Min	Max	SD	No	Mean	Min	Max	SD
Cast. Temp. °C	147	1237	1215	1256	11.40	96	1236	1210	1270	11.82
C	155	3.96	3.67	4.27	0.087	99	3.91	3.74	4.06	0.073
Si	155	1.52	1.27	1.87	0.106	99	1.47	1.09	1.71	0.113
CEV	155	4.47	4.20	4.89	0.101	99	4.40	4.20	4.56	0.087
S	153	0.044	0.019	0.070	0.010	99	0.048	0.022	0.079	0.013
P	37	0.048	0.035	0.064	0.007	18	0.049	0.042	0.061	0.005
Mn	155	0.75	0.57	0.99	0.063	99	0.75	0.64	0.87	0.043
Ti	152	0.040	0.010	0.113	0.014	95	0.030	0.011	0.061	0.010
USV, km/s	137	3.54	3.25	4.00	0.138	63	3.65	3.24	4.10	0.164
Variable	63/17/20					63/17/20+Ti				
	No	Mean	Min	Max	SD	No	Mean	Min	Max	SD
Cast. Temp.	67	1250	1220	1280	14.06	47	1254	1230	1290	11.83
C	65	3.92	3.76	4.09	0.066	49	3.92	3.78	4.22	0.089
Si	65	1.50	1.25	1.69	0.099	49	1.50	1.27	1.98	0.153
CEV	65	4.42	4.21	4.64	0.077	49	4.42	4.26	4.78	0.104
S	65	0.046	0.025	0.072	0.011	49	0.047	0.013	0.069	0.013
P	0					2	0.058	0.058	0.058	
Mn	64	0.76	0.60	0.89	0.054	49	0.73	0.53	0.83	0.061
Ti	64	0.040	0.019	0.066	0.009	48	0.051	0.020	0.118	0.015
USV, km/s	8	3.59	3.32	3.95	0.237	20	3.46	3.23	3.80	0.147

Table 33 t-Tests For Significance Between Burden Groups For Fullwood 48-Type Moulds

Variable		70/20/10 v 53/17/30	53/17/30 v 63/17/20	63/17/20 v 63/17/20+Ti	70/20/10 v 63/17/20	70/20/10 v 63/17/20+Ti	53/17/30 v 63/17/20+Ti
Casting Temp. °C	No.	241	161	112	212	192	141
	Comb. σ	11.57	12.79	13.19	12.29	11.56	11.82
	t	0.66	6.88	1.59	7.19	8.82	8.55
	Significance	NS	99.9%	NS	99.9%	99.9%	99.9%
C	No.	252	162	112	218	202	146
	Comb. σ	0.0818	0.0703	0.0764	0.0812	0.0874	0.0787
	t	5.22	1.43	0.55	3.25	3.28	0.58
	Significance	99.9%	NS	NS	99%	99%	NS
Si	No.	252	162	112	218	202	146
	Comb. σ	0.109	0.108	0.125	0.104	0.119	0.127
	t	3.39	1.51	0.04	1.37	1.03	1.21
	Significance	99.9%	NS	NS	NS	NS	NS
CEV	No.	252	162	112	218	202	146
	Comb. σ	0.0956	0.0831	0.0893	0.0942	0.102	0.0929
	t	5.77	1.81	0.41	3.37	3.25	1.05
	Significance	99.9%	NS	NS	99.9%	99%	NS
S	No.	250	162	112	216	200	146
	Comb. σ	0.0111	0.0119	0.0116	0.0102	0.0107	0.0127
	t	2.59	0.63	0.27	1.66	1.76	0.27
	Significance	99%	NS	NS	NS	NS	NS
P	No.	53	-	-	-	37	18
	Comb. σ	0.0066	-	-	-	0.0071	0.0049
	t	0.78	-	-	-	2.00	2.49
	Significance	NS	-	-	-	NS	95%

Cont'd

Table 33 (Cont'd)

Variable		70/20/10 v 53/17/30	53/17/30 v 63/17/20	63/17/20 v 63/17/20+Ti	70/20/10 v 63/17/20	70/20/10 v 63/17/20+Ti	53/17/30 v 63/17/20+Ti
Mn	No.	252	161	111	217	202	146
	Comb. σ	0.0562	0.0477	0.0569	0.0604	0.0625	0.0498
	t	0.19	0.68	2.15	0.42	1.89	2.07
	Significance	NS	NS	95%	NS	NS	95%
Ti	No.	245	157	110	214	198	141
	Comb. σ	0.0125	0.0098	0.0118	0.0124	0.0139	0.0120
	t	5.64	6.10	5.02	0.27	5.13	9.84
	Significance	99.9%	99.9%	99.9%	NS	99.9%	99.9%
USV	No.	198	69	26	143	155	81
	Comb. σ	0.147	0.173	0.176	0.145	0.139	0.161
	t	4.87	1.05	1.71	0.78	2.55	4.71
	Significance	99.9%	NS	NS	NS	95%	99.9%

Table 34 Summary Of Significant Production And Life Changes For 48-Type Moulds

Variable	Significant Increase			
	70/20/10	53/17/20/10	63/17/20	63/17/20 + Ti
Cast T°C	1237 → 1250	1236 → 1250		
	1237 → 1254			
		1236 → 1254		
S	0.044 → 0.048			
P		0.049 → 0.058		
Ti		0.030 → 0.040		
			0.040 → 0.051	
	0.040 → 0.051			
		0.030 → 0.051		
USV	3.54 → 3.65			
LIFE		54.94 → 63.34		

Variable	Significant Decrease			
	70/20/10	53/17/20/10	63/17/20	63/17/20 + Ti
C	3.96 → 3.91			
	3.96 → 3.92			
	3.96 → 3.92			
Si	1.52 → 1.47			
CEV	4.46 → 4.40			
	4.46 → 4.42			
	4.46 → 4.42			
Mn			0.76 → 0.73	
		0.75 → 0.73		
Ti	0.040 → 0.030			
USV	3.54 → 3.46			
		3.65 → 3.46		
LIFE	67.88 → 54.94			

Table 35 Nitrogen Analyses Of Fullwood Ingot Moulds

Burden	Sample Type	Mould No.	Nitrogen,%
70/20/10	Lug	180/461	0.008
		181/965	0.018
		48/695	0.009
		48/723	0.008
		49/884	0.012
		49/885	0.006
		49/887	0.008
53/17/20/10	Lug	48/778	0.010
		48/779	0.011
		48/780	0.008
		48/803	0.009
		49/912	0.010
		49/930	0.009
		26/846	0.012
		26/848	0.0065
		26/849	0.007
		43/977	0.009
		44/161	0.006
		45/030	0.006
		45/034	0.005
		46/875	0.006
		46/876	0.007
		46/880	0.007
		46/883	0.010
		49/913	0.008
		49/921	0.009
		49/922	0.009
		48/842	0.0055
		26/838	0.006
		26/843	0.0059
		26/830	0.0068
		46/768	0.0056
53/17/20/10	Lug	48/791	0.0050
		48/789	0.0073
		26/854	0.0051
		44/173	0.0049
	Ladle	953/768	0.0081
		28/165	0.0090
		953/790	0.0091
		953/792	0.0082
		48/774	0.0103
		953/793	0.0072
		161/350	0.0080
		48/776	0.0081
		48/783	0.0072
		48/790	0.0086
		48/793	0.0073
		48/794	0.0086
		49/911	0.0097
		49/912	0.0070
		48/811	0.0079
63/17/20	Ladle	46/932	0.0071
		48/879	0.0071
		161/371	0.0095
		49/964	0.0099
		49/964	0.0101
		48/880	0.0087
		160/110	0.0095
		181/057	0.0077
		057/411	0.0084
		057/411	0.0085
		46/933	0.0075
		03/1/67	0.0106

Table 36 Trace Element Levels In Fullwood Iron

Burden	Mould No.	Mo	Cr	Sn	V	Ni	Al	Cu
70/20/10	180/461	0.002	0.01	0.004	0.11	0.04	0.007	0.06
	181/965	0.002	0.01	0.004	0.06	0.03	0.022	0.04
	48/695	0.005	0.01	0.004	0.04	0.02	0.011	0.04
	48/723	0.002	0.01	0.004	0.01	0.02	0.010	0.04
	49/884	0.002	0.01	0.004	0.01	0.03	0.014	0.05
	49/885	0.003	0.02	0.004	0.10	0.05	0.014	0.04
	49/887	0.002	0.01	0.004	0.08	0.04	0.012	0.06
53/17/20/10	48/778	0.003	0.01	0.004	0.01	0.03	0.014	0.08
	48/779	0.006	0.03	0.004	0.01	0.05	0.011	0.06
	48/780	0.004	0.02	0.004	0.01	0.03	0.009	0.09
	48/803	0.003	0.01	0.004	<0.01	0.03	0.010	0.06
	49/912	0.003	0.01	0.004	0.01	0.03	0.008	0.08
	49/930	0.004	0.02	0.004	<0.01	0.02	0.007	0.04
	26/846	0.003	0.01	0.004	<0.01	0.02	0.007	0.04
	26/848	0.004	0.03	0.005	<0.01	0.03	0.009	0.06
	26/849	0.004	0.02	0.004	<0.01	0.04	0.010	0.06
	48/774	-	0.22	0.002	0.03	0.02	-	0.04
	48/771	<0.01	0.02	0.002	0.02	0.01	0.009	0.04
	48/773	<0.01	0.01	0.002	0.01	0.02	0.006	0.06
	48/775	<0.01	0.01	0.002	0.01	<0.01	0.007	0.03
	48/776	<0.01	0.01	-	0.01	0.02	0.009	0.05
	48/792	<0.01	0.01	0.002	0.01	0.01	0.008	0.04
	48/805	<0.01	0.02	0.005	0.01	0.02	0.003	0.08
	43/997	0.003	0.01	0.004	0.02	0.03	0.007	0.04
	44/161	0.004	0.01	0.004	<0.01	0.03	0.007	0.07
	45/030	0.003	0.01	0.004	0.01	0.03	0.006	0.07
	45/034	0.004	0.02	0.004	<0.01	0.03	0.008	0.06
	46/875	0.003	0.01	0.004	<0.01	0.02	0.007	0.05
	46/876	0.003	0.01	0.004	0.02	0.02	0.009	0.05
	46/880	0.002	0.02	0.004	<0.01	0.02	0.007	0.05

Table 37 Bend Test Properties Of 48-Type Moulds

Burden	Graphite Form	Mould No.	Life	Mode Of Failure	Max. Load kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A/B	A	B
70/20/10	Flake	48/585	120	VC/BS	1.0	0.5	0.6	0.3	0.6	1.7
53/17/20/10	Flake	48/793	24	VC/BS	1.2	0.9	1.2	0.5	0.4	0.8
53/17/20/10	Flake	48/839	15	VC/NS	1.6	2.5	3.7	1.4	0.8	0.5
63/17/20+Ti	Flake	48/059	2	Torn Seat	1.3	1.7	2.1	0.9	1.0	1.1
Distington	Compacted	48/758	65	VC/BS	2.4	2.2	2.7	1.3	3.7	2.9
53/17/20/10	Compacted	48/840	16	VC/NS	2.6	4.2	5.2	2.0	1.6	0.8
63/17/20	Compacted	48/879	13	VC/C	2.6	2.5	3.5	1.7	3.6	2.0

Table 38 Summary Of Mould Production Data For River Don Forging Moulds

Cast Report No.	Cast Date	Mould Type	No.	C	Si	S	P	Mn	Ti	USV km/s	Comments
59	19.11.76	71/68	1	3.85	1.90	0.015	0.048	0.70	0.045	3.52	Heavy crust on Ladle. Not Purged. Little sand on rising surfac
111	26.5.77	79/75	1	3.80	1.65	0.002	0.050	0.76	0.07		Purged 9 mins. Ran to slag. $\frac{1}{4}$t back feed.
131	13.10.77	105/101	1	3.95	2.02	0.017	0.116	0.72	0.065	3.34	Purged 23 mins. Backed up.
143	13.12.77	91/86	1	3.85	1.95	0.007	0.057	0.80	0.055	3.46	Not purged.
144	15.12.77	105/101	2	3.92	1.85	0.009	0.083	0.80	0.055	3.36	Not purged.
161	16.5.78	105/101	3	3.86	2.00	0.012	0.065	0.83	0.055	3.40	Purged. Inoculated. Ran to slag.
180	15.9.78	71/68 x 16	1	3.70	1.90	0.009	0.050	0.70	0.045	3.30	Purged 30 mins.
189	1.11.78	91/86	2	3.76	1.55	0.008	0.05	0.72	0.04	3.46	Purged 2 mins. Slow casting rate.
210	23.3.79	105/101	4	3.80	1.50	0.012	0.050	0.70	0.045	3.44	Whirlgates.
213	3.4.79	71/68 x 16	2	3.90	1.62	0.052	0.048	0.80	0.045		Purged 77 mins. Whirlgates.
221	24.5.79	105/101	5	3.83	1.56	0.034	0.059	0.71	0.022	3.80	Not purged. Whirl-gates.

Table 39 Summary Of Mould Usage Data For River Don Forging Moulds

Cast Report No.	Cast Date	Mould Type	No.	Current Life	Comments
59	19.11.76	71/68	1	29	O.K.
111	26.5.77	79/75	1	9	One panel plucked after 2nd life.Weld repaired.
131	13.10.77	105/101	1	9	Roughening on 6 panels after 3rd life.Ground clear. 8 panels plucked after 8th life.
143	13.12.77	91/86	1	10	8 panels plucked after 9th life.Weld repaired.
144	15.12.77	105/101	2	9	3 panels plucked after 2nd life.2panels welded and 3rd ground clear.5 panels plucked after 5th cast. Welded.8 panels plucked after 8th cast.Major weld repair.1 panel plucked after 9th life
161	16.5.78	105/101	3	8	Slight plucking 4 faces after 2nd cast.Cracked and plucked after 3rd cast.Badly plucked after 8th life.
180	15.9.78	71/68x16	1	4	O.K.
189	1.11.78	91/86	2	2	O.K.
210	23.3.79	105/101	4	2	Cracked after 1st life. Stitched.
213	3.4.79	71/68x16	2	1	O.K.
221	24.5.79	105/101	5	1	O.K.

Table 40 Chemical Analyses Of Renishaw F Moulds From January,1973 To January,1975

Variable	No. Of Moulds	Mean	Min.	Max.	Standard Deviation
Carbon	402	3.79	3.50	4.21	0.099
Silicon	402	1.23	0.68	1.86	0.174
Carbon equivalent	402	4.20	3.95	4.76	0.121
Manganese	402	0.93	0.23	1.42	0.123
Sulphur	402	0.080	0.024	0.175	0.025
Phosphorus	402	0.063	0.026	0.264	0.028
Molybdenum	377	0.016	0.004	0.068	0.007
Chromium	377	0.080	0.001	0.39	0.035
Tin	377	0.006	0.001	0.090	0.006
Vanadium	377	0.016	0.001	0.105	0.012
Titanium	377	0.018	0.001	0.112	0.022
Nickel	377	0.057	0.001	0.192	0.024
Aluminium	377	0.002	0.002	0.008	0.001
Copper	377	0.106	0.025	0.38	0.044
Performance	402	116.7	18	270	31.9

Table 41 Monthly Average Performance And Phosphorus Content Of Renishaw F Type Moulds

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
January 1973	19	116.3	63	155	20.1	0.058	0.043	0.081	0.009
February 1973	18	127.9	64	170	26.9	0.057	0.028	0.073	0.011
March 1973	16	110.9	72	131	19.6	0.052	0.026	0.071	0.014
April 1973	13	115.8	64	151	27.9	0.057	0.041	0.072	0.011
May 1973	17	124.4	68	173	29.6	0.057	0.030	0.077	0.016
June 1973	29	120.9	24	182	36.4	0.052	0.027	0.076	0.014
July 1973	29	133.2	70	174	20.6	0.051	0.029	0.072	0.010
August 1973	25	118.8	19	151	28.8	0.059	0.036	0.101	0.013
September 1973	28	116.7	60	170	28.0	0.053	0.033	0.065	0.008
October 1973	22	127.0	74	157	27.8	0.065	0.039	0.180	0.028
November 1973	29	113.8	40	164	26.3	0.052	0.039	0.067	0.007
December 1973	10	122.9	74	159	39.9	0.051	0.039	0.057	0.006
January 1974	10	121.4	94	154	22.4	0.057	0.051	0.073	0.007
		No F Moulds Produced Between Feb. To July 1974							
August 1974	3	59.3	43	78	17.6	0.182	0.177	0.190	0.007
September 1974	8	78.8	41	134	32.1	0.117	0.035	0.264	0.087
October 1974	15	87.1	30	155	40.1	0.125	0.039	0.228	0.051
November 1974	9	75.4	29	119	34.4	0.119	0.075	0.165	0.027
December 1974	11	92.8	18	133	41.5	0.079	0.052	0.169	0.040
January 1974	4	110.8	70	140	29.5	0.054	0.039	0.065	0.011

Table 42 The Performance Of Renishaw F Type Moulds With Respect To Phosphorus Content-All Moulds Produced Over Period January 1973 To January 1975

Phosphorus Content %	No. Of Moulds	Performance			
		Mean	Min.	Max.	S.D.
Less Than 0.07	327	120.2	19	182	26.5
0.071 to 0.13	50	119.9	70	193	24.5
0.131 to 0.20	22*	68.5	18	177	26.3
More Than 0.2	2*	62	60	64	

* Includes Heat Treated Moulds

Table 43 Effect On Life Of Moulds Cast On Fridays For Renishaw F Type Moulds Cast August To December 1974

Day Cast	≤0.12%P			>0.12%P		
	No. Of Moulds	Mean Life	S.D.	No. Of Moulds	Mean Life	S.D.
Monday To Thursday	25	109.9	31.1	17	49.4	31.4
Friday	6	131.7	20.8	9	48.4	21.2

Table 44 Chemical Analyses Of Renishaw 610 WB Moulds From January 1973 To February 1975

Variable	No. Of Moulds	Mean	Min	Max	Standard Deviation
Carbon	558	3.83	3.45	4.19	0.10
Silicon	558	1.20	0.77	1.89	0.16
Carbon Equivalent	558	4.22	3.83	4.61	0.115
Manganese	558	0.95	0.21	1.20	0.09
Sulphur	558	0.079	0.004	0.155	0.021
Phosphorus	558	0.089	0.025	0.319	0.053
Molybdenum	558	0.012	0.001	0.100	0.006
Chromium	558	0.067	0.002	0.160	0.020
Tin	558	0.004	0.001	0.088	0.005
Vanadium	558	0.006	0.001	0.080	0.008
Titanium	558	0.003	0.001	0.063	0.005
Nickel	558	0.052	0.008	0.137	0.019
Aluminium	558	0.002	0.001	0.112	0.006
Copper	558	0.082	0.010	0.300	0.035
Performance	558	82.6	3	173	23.8

Table 45 The Monthly Avaerage Performance And Phosphorus Content Of Renishaw 610 WB Moulds

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
January 1973	3	98.3	95	102	12.3	0.068	0.055	0.083	0.014
February 1973	37	98.1	69	134	16.6	0.057	0.028	0.074	0.013
March 1973	93	96.1	59	177	19.7	0.056	0.024	0.082	0.012
April 1973	74	91.4	24	124	24.2	0.061	0.038	0.098	0.010
May 1973	52	100.6	29	135	19.3	0.058	0.023	0.075	0.013
June 1973	86	100.5	40	147	19.5	0.061	0.027	0.081	0.011
July 1973	58	94.6	40	180	24.6	0.051	0.026	0.072	0.010
August 1973	36	91.1	31	132	22.4	0.056	0.042	0.081	0.008
September 1973	48	93.4	30	173	20.4	0.056	0.042	0.074	0.010
October 1973	79	90.8	51	164	19.0	0.062	0.043	0.137	0.016
November 1973	30	85.4	42	101	15.1	0.050	0.026	0.068	0.007
December 1973	12	97.0	59	124	19.7	0.059	0.045	0.085	0.013
January 1974	4	104.8	96	126	14.3	0.053	0.048	0.056	0.004
March 1974	8	105.4	83	123	13.5	0.055	0.048	0.067	0.008
April 1974	24	93.2	55	141	20.1	0.079	0.042	0.297	0.053
May 1974	27	94.7	69	134	16.6	0.062	0.044	0.123	0.016
June 1974	58	87.7	15	138	22.8	0.066	0.032	0.319	0.051
July 1974	37	97.9	35	141	27.5	0.111	0.044	0.302	0.072
August 1974	54	88.1	48	133	20.9	0.124	0.040	0.248	0.068

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Table 45 Cont'd

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
September 1974	46	94.6	64	131	15.7	0.095	0.045	0.214	0.053
October 1974	65	83.1	43	124	17.9	0.129	0.040	0.239	0.053
November 1974	63	59.6	3	111	23.2	0.072	0.046	0.145	0.024
January 1975	62	79.6	10	110	18.5	0.061	0.025	0.083	0.011
February 1975	65	84.4	36	114	19.8	0.061	0.033	0.105	0.016

Table 46 The Performance Of Renishaw 610 WB moulds With Respect To Phosphorus Content

Phosphorus Content %	No. Of Moulds	Performance			
		Mean	Min.	Max.	S. D.
Less Than 0.07	318	87.3	13	172	23.5
0.071 to 0.13	100	74.5	5	141	24.2
0.131 to 0.20	111	77.4	3	135	22.7
More than 0.20	15	78.9	44	110	19.1

Table 47 The Mean Phosphorus Levels At Various Life Ranges For Renishaw 610 WB moulds

	Performance Range						
	0-10	11-20	21-30	31-40	41-50	51-60	61-70
Mean Phos. %	0.145	0.102	0.058	0.087	0.092	0.087	0.074

Table 48 Chemical Analyses Of Renishaw B120 Moulds Manufactured Between April 1972 And December 1974

Variable	No. Of Moulds	Mean	Min	Max	Standard Deviation
Carbon	581	3.81	3.55	4.15	0.089
Silicon	581	1.22	0.78	1.82	0.172
Carbon Equivalent	581	4.22	3.85	4.61	0.109
Manganese	581	0.92	0.23	1.18	0.114
Sulphur	581	0.078	0.005	0.166	0.024
Phosphorus	581	0.071	0.016	0.255	0.036
Molybdenum	320	0.014	0.002	0.060	0.006
Chromium	320	0.070	0.012	0.180	0.021
Tin	320	0.005	0.001	0.022	0.003
Vanadium	320	0.009	0.001	0.044	0.009
Titanium	320	0.004	0.001	0.060	0.005
Nickel	320	0.054	0.001	0.120	0.019
Aluminium	320	0.002	0.001	0.016	0.001
Copper	320	0.090	0.032	0.30	0.032
Performance	581	106.5	12	175	22.9

Table 49 The Monthly Average Performance And Phosphorus Content Of Renishaw B120 Moulds

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
April 1972	2	98.5	66	131	45.9	0.064	0.061	0.067	0.004
May 1972	5	121.4	75	154	29.7	0.060	0.045	0.084	0.014
July 1972	2	106.5	69	144	53.2	0.070	0.069	0.071	0.001
August 1972	3	94.7	68	140	39.5	0.068	0.062	0.086	0.016
October 1972	4	94.8	89	100	4.8	0.061	0.050	0.073	0.009
November 1972	24	100.5	78	141	14.0	0.006	0.046	0.085	0.009
December 1972	44	97.7	55	172	19.7	0.061	0.016	0.076	0.011
January 1973	16	94.9	59	175	19.4	0.059	0.040	0.074	0.009
February 1973	25	96.2	66	152	24.3	0.062	0.023	0.080	0.011
March 1973	15	110.3	73	136	20.3	0.059	0.020	0.075	0.013
April 1973	9	135.5	107	159	18.6	0.059	0.042	0.075	0.011
May 1973	4	138.0	119	175	25.2	0.048	0.027	0.061	0.014
June 1973	13	114.7	73	158	25.7	0.062	0.036	0.082	0.012
July 1973	7	124.3	106	164	20.3	0.051	0.040	0.066	0.008
August 1973	12	123.5	109	141	8.3	0.058	0.040	0.079	0.011
September 1973	6	126.5	108	153	18.8	0.057	0.049	0.067	0.006
November 1973	5	111.8	90	155	25.8	0.050	0.045	0.057	0.005
December 1973	11	86.3	58	102	12.8	0.055	0.034	0.067	0.008
January 1974	40	108.7	85	145	13.4	0.057	0.039	0.076	0.008

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Table 49 Cont'd

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
February 1974	50	122.7	65	155	20.4	0.064	0.037	0.104	0.015
March 1974	47	108.1	37	154	25.4	0.070	0.047	0.153	0.025
April 1974	14	110.9	83	157	21.4	0.070	0.028	0.244	0.053
June 1974	12	126.5	106	146	12.4	0.051	0.041	0.063	0.006
July 1974	16	116.3	19	151	32.0	0.112	0.040	0.242	0.072
August 1974	17	114.5	90	140	13.9	0.128	0.043	0.244	0.067
September 1974	28	111.5	58	155	27.4	0.123	0.038	0.255	0.077
October 1974		100.5	49	132	19.9	0.107	0.042	0.157	0.048
November 1974	30	100.0	73	134	14.1	0.099	0.010	0.152	0.036
December 1974	23	95.7	12	116	20.3	0.071	0.045	0.176	0.033

Table 50 The Performance Of Renishaw B120 Moulds With Respect To Phosphorus Content

Phosphorus Content %	No. Of Moulds	Performance			
		Mean	Min.	Max.	S.D.
Less Than 0.07	442	107.3	42	158	22.0
0.071 to 0.13	86	102.2	12	154	25.7
0.131 to 0.20	55	103.2	12	151	25.3
More Than 0.2	9	103.3	65	150	29.4

Table 51 The Mean Phosphorus Contents Of Renishaw B120 Moulds With Respect To Performance

	Performance Range				
	0-30	31-40	41-50	51-60	61-70
Mean Phos. %	0.169	0.077	0.052	0.079	0.073

Table 52 Chemical Analyses Of Renishaw WEU 100 Moulds Scrapped Between September 1974 And March 1976

Variable	No. Of Moulds	Mean	Min	Max	Standard Deviation
Carbon	336	3.81	3.48	4.07	0.097
Silicon	336	1.21	0.98	1.89	0.167
Sulphur	336	0.075	0.012	0.129	0.019
Phosphorus	336	0.074	0.030	0.329	0.046
Manganese	336	0.95	0.58	1.30	0.090
Molybdenum	336	0.012	0.001	0.065	0.006
Chromium	336	0.067	0.007	0.30	0.020
Tin	336	0.004	0.001	0.015	0.003
Vanadium	336	0.011	0.001	0.071	0.011
Titanium	336	0.005	0.001	0.078	0.007
Nickel	336	0.051	0.001	0.150	0.020
Aluminium	336	0.002	0.001	0.008	0.001
Copper	336	0.084	0.021	0.230	0.031
Performance	336	62.9	6	113	18.77

Table 53 The Monthly Average Performance And Phosphorus Content Of Renishaw WEU100 Moulds Scrapped Between September 1974 And March 1976

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
July 1973	1	57				0.054			
September 1973	1	69				0.061			
November 1973	3	70.0	59	88	15.7	0.051	0.043	0.062	0.010
December 1973	2	83.5	83	84	0.7	0.067	0.063	0.071	0.006
January 1974	5	76.4	70	90	8.2	0.056	0.053	0.065	0.005
February 1974	9	73.0	47	113	19.7	0.070	0.050	0.095	0.017
March 1974	20	76.2	47	96	14.3	0.065	0.046	0.109	0.016
April 1974	18	60.6	6	100	21.3	0.104	0.042	0.295	0.077
May 1974	20	67.6	49	97	12.4	0.064	0.043	0.200	0.033
June 1974	20	69.8	36	109	18.3	0.062	0.040	0.193	0.032
July 1974	16	67.7	29	113	21.7	0.142	0.042	0.329	0.093
August 1974	11	59.4	11	80	18.3	0.117	0.042	0.244	0.064
September 1974	25	59.1	22	80	13.6	0.112	0.041	0.287	0.070
October 1974	8	61.3	28	83	18.8	0.102	0.040	0.166	0.042
November 1974	7	44.3	20	85	25.8	0.131	0.108	0.151	0.017
December 1974	3	62.0	48	74	13.1	0.052	0.047	0.058	0.006
January 1975	25	66.4	16	98	19.0	0.065	0.039	0.132	0.018
February 1975	22	58.5	32	92	17.2	0.062	0.037	0.100	0.017

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Table 53 Cont'd

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
March 1975	18	56.7	22	75	15.8	0.044	0.030	0.078	0.011
April 1975	20	58.7	19	95	17.8	0.059	0.041	0.092	0.011
May 1975	16	58.9	17	86	18.5	0.059	0.043	0.075	0.009
June 1975	20	55.8	8	102	21.4	0.059	0.043	0.083	0.012
July 1975	9	60.5	20	97	22.8	0.048	0.044	0.051	0.003
August 1975	3	52.3	37	66	14.6	0.067	0.060	0.074	0.007
September 1975	18	66.0	46	86	11.4	0.056	0.045	0.072	0.007
October 1975	11	66.7	35	98	21.4	0.052	0.037	0.060	0.008
November 1975	4	64.0	58	79	10.0	0.057	0.046	0.069	0.011
December 1975	1	55				0.044			

Table 54 The Performance Of Renishaw WEU100 Moulds With Respect To Phosphorus Content

Phosphorus Content %	No. Of Moulds	Performance			
		Mean	Min.	Max.	S.D.
Less Than 0.07	244	64.96	16	113	17.43
0.071 to 0.13	53	60.23	6	96	20.57
0.131 to 0.20	29	59.90	11	102	19.60
More Than 0.2	10	57.40	29	78	16.00

Table 55 Mean Phosphorus Levels At Various Life Ranges For Renishaw WEU100 Moulds

	Performance Range						
	0-30	31-40	41-50	51-60	61-70	71-80	81-90
Mean Phos. %	0.102	0.100	0.070	0.078	0.074	0.065	0.064

Table 56 Chemical Analyses Of Renishaw NEU 100 Moulds Scrapped Between September 1974 And March 1976

Variable	No. Of Moulds	Mean	Min	Max	Standard Deviation
Carbon	159	3.81	3.60	4.07	0.093
Silicon	159	1.24	0.96	1.69	0.162
Sulphur	159	0.075	0.070	0.174	0.024
Phosphorus	159	0.066	0.031	0.263	0.035
Manganese	159	0.95	0.59	1.10	0.093
Molybdenum	159	0.012	0.010	0.032	0.005
Chromium	159	0.067	0.010	0.180	0.023
Tin	159	0.005	0.001	0.040	0.004
Vanadium	159	0.009	0.001	0.039	0.009
Titanium	159	0.005	0.001	0.067	0.007
Nickel	159	0.052	0.001	0.152	0.023
Aluminium	159	0.002	0.001	0.010	0.002
Copper	159	0.087	0.016	0.187	0.033
Performance	159	95.8	16	149	24.6

Table 57 Chemical Analyses Of Renishaw NEU100 Moulds Scrapped Between September 1974 And March 1976

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
February 1973	1	82				0.043			
March 1973	2	138.0	127	149	15.6	0.058	0,050	0.066	0.011
May 1973	1	102				0.066			
June 1973	1	112				0.056			
July 1973	3	99.7	87	122	19.4	0.041	0.031	0.050	0.010
August 1973	1	82				0.046			
September 1973	3	91.3	36	126	48.43	0.049	0.046	0.053	0.004
October 1973	5	99.2	83	115	12.01	0.058	0.047	0.074	0.011
November 1973	2	68.5	57	80	16.3	0.048	0.045	0.050	0.004
December 1973	4	103.8	92	122	14.2	0.052	0.047	0.058	0.005
January 1974	5	90.6	70	125	21.9	0.053	0.046	0.059	0.005
February 1974	5	88.2	58	112	22.7	0.067	0.056	0.079	0.009
March 1974	7	113.6	90	132	13.3	0.066	0.050	0.099	0.017
May 1974	4	91.5	70	112	18.4	0.059	0.045	0.076	0.013
June 1974	7	72.1	16	125	43.0	0.046	0.035	0.056	0.008
July 1974	4	114.8	107	127	8.6	0.167	0.111	0.263	0.067
August 1974	7	95.3	76	126	16.5	0.136	0.040	0.208	0.067
September 1974	2	97.5	86	109	16.3	0.114	0.042	0.185	0.100

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Table 57 Cont'd

Month Cast	No. Of Moulds	Performance				Phosphorus			
		Mean	Min	Max	SD	Mean	Min	Max	SD
October 1974	5	95.0	77	146	29.6	0.129	0.058	0.202	0.053
December 1974	9	81.7	16	125	39.6	0.066	0.055	0.096	0.013
January 1975	14	99.6	56	138	23.5	0.064	0.045	0.084	0.010
February 1975	14	106.9	65	135	20.8	0.065	0.034	0.096	0.019
March 1975	18	100.8	81	142	17.4	0.049	0.031	0.073	0.011
April 1975	11	88.2	40	123	23.9	0.052	0.038	0.068	0.009
May 1975	13	97.7	49	122	22.6	0.057	0.044	0.075	0.007
June 1975	7	83.6	61	109	17.6	0.055	0.045	0.072	0.010
July 1975	2	88.0	75	101	18.3	0.053	0.052	0.053	0.001

Table 58 The Performance Of Renishaw NEU100 Moulds With Respect To Phosphorus Content

Phosphorus Content %	No. Of Moulds	Performance			
		Mean	Min.	Max.	S.D.
Less than 0.07	127	95.2	16	149	24.28
0.071 to 0.13	23	94.7	16	131	27.70
0.131 to 0.2	8	109.9	76	146	23.69
More than 0.2	3	99.0	89	112	11.79

Table 59 Mean Phosphorus Levels At Various Life Ranges For Renishaw NEU100 Moulds

	Performance Range						
	0-30	31-40	41-50	51-60	61-70	71-80	81-90
Mean Phos. %	0.073	0.051	0.044	0.063	0.070	0.072	0.060
	Performance Range						
	91-100	101-110	111-120	121-130	131-140		
Mean Phos. %	0.059	0.068	0.072	0.071	0.056		

Table 60 Stripping Details At Renishaw- Moulds Cast On October 13th 1977

Mould Type	No.	Time Cast	Time Top Plate Removed	Sand Condition After Top Plate Removal	Top Temp °C	Cast To Remove Top, h	Cast To Drop Barrel & Strip Box, h	Condition Of Sand Jacket After Box Removal
Atlas (2.7t)	3456	1615	2303	Up to 50mm on ¾ of top	765	6¾	8	O.K.
"	3455	1550	2305	"	760	7¼	8¼	O.K.
"	3454	1400	2306	"	660	9	10	O.K.
"	3453	1135	2307	"	675	11½	12½	Small part of 1 corner exposed
FB 7ACX (2.1t)	77	1545	2309	"	720	7½	8¾	Nearly all sand off. Temp. 800°C.
Brymbo(2.9t)	1212	1555	2310	"	630	7¼	8½	¼ sand off. 780°C
Patent Shaft (3.1t)	1009	1610	2311	None	580	7	8½	¾ sand off. 670°C
610WB (5.0t)	3629	1400	2312	"	760	9¼	10¾	O.K.
"	3630	1130	2313	"	630	11¾	13¼	Half sand thickness fall off-lugs and part of corner exposed.
F (6.1t)	3824	1315	2315	"	960	10	12	25-75mm left on ½ mould
"	3823	1100	2318	"	960	12¼	14¼	1 panel fallen off.
WEU (5.5t)	687	1000	2320	"	500	13¼	15¼	"

Table 61 Stripping Details At Renishaw- Moulds Cast On October 20th 1977

Mould Type	No.	Time Cast	Time Top Plate Removed	Sand Condition After Top Plate Removal	Top Temp °C	Cast To Remove Top, h	Cast To Drop Barrel & Strip Box, h	Condition Of Sand Jacke After Box Removal
Atlas (2.7t)	3476	1400	2349	Up to 50mm on ¾ of top	660	9¾	10	O.K.
"	3408	1310	2350	"	645	10¾	11	2 small parts of bottom corners exposed
"	3478	1275	2353	"	650	11½	11¾	O.K.
"	3479	1035	2353	"	620	13¼	13¾	O.K.
"	3477	1440	2356	"	500	9¼	9¾	O.K.
FB 7ACX (2.1t)	82	1655	2355	"	650	9	9¾	Only ½ remaining-760°C
Brymbo(2.9t)	1222	1650	2354	"	730	9	9¾	None remaining-800°C
ESCF (5.7t)	3902	1115	2335	None	890	12¼	14¾	Thin on ¼ mould
WEU (5.5t)	692	1220	2346	"	570	11½	13½	O.K.
RDG (7.0t)	530	1015	2347	"	540	13½	15½	O.K.
ESCF (5.7t)	3901	1150	2348	"	520	13	15	Thin on 1 side
610WB (5.0t)	3639	1305	2332	"	610	10½	12½	Thin on 2 sides
"	3640	1355	2331	"	620	9½	11½	Thin in ½ of 1 side
"	3641	1435	2330	"	650	9	11	Thin on 1 side

Table 62 Stripping Details At Renishaw- Moulds Cast On October 27th 1977

Mould Type	No.	Time Cast	Time Top Plate Removed	Sand Condition After Top Plate Removal	Top Temp °C	Cast To Remove Top, h	Cast To Drop Barrel & Strip Box, h	Condition Of Sand Jacket After Box Removal
Atlas(2.7t)	3501	1220	2301	Up to 50 mm on $\frac{3}{4}$ of top.	700	$10\frac{3}{4}$	12	O.K.
Atlas(2.7t)	3499	1145	2302	"	700	$11\frac{3}{4}$	12	O.K.
Atlas(2.7t)	3500	1115	2305	"	730	$11\frac{3}{4}$	$12\frac{1}{4}$	Small part of bottom expos
H.W.	12	1255	_	Top still on.			$10\frac{1}{2}$	O.K.
Brymbo(2.9t)	1184	1635	2300	Adherent covering (CO_2)		$6\frac{1}{2}$	$7\frac{3}{4}$	2 panels exposed-surface temp. 890°C.
FB 7ACX(2.1t)	87	1630	-	Top still on.			8	Thin on two faces.
15" tyre mould.	8	1140	-	"			$13\frac{1}{4}$	No sand remaining-surface temp. 830°C.
B125(6.4t)	1894	0410	2311	None.	700	19	$20\frac{1}{2}$	Thin on one corner.
B125(6.4t)	1903	1335	2313	None.	750	$9\frac{3}{4}$	$11\frac{1}{2}$	Thin on two faces.
SPT(5.0t)	3653	1255	2317	None.	780	$10\frac{1}{4}$	$11\frac{3}{4}$	Thin on one face.
SPT(5.0t)	3654	1215	2319	None.	710	11	$12\frac{3}{4}$	Thin on three faces.
SPT(5.0t)	3652	1110	2324	None.	760	$12\frac{1}{4}$	14	O.K.
ESCF(5.7t)	3910	0905	2321	None.	880	$14\frac{1}{4}$	$16\frac{1}{4}$	Thin on three faces.
ESCF(5.7t)	3911	1025	2322	None.	910	13	15	Thin on one side.
WEU(5.5t)	697	0940	2320	None.	770	$13\frac{3}{4}$	$15\frac{3}{4}$	Thin on two faces.

Table 63 Analyses Of WEU100 Moulds

Mould	Life	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Al	Cu	Calc. Life
404	30	3.79	1.29	.099	.053	.87	.022	.096	.013	.057	.012	.055	.002	.057	51±42
489	6	3.65	1.36	.069	.052	.79	.020	.051	.025	.011	.009	.052	.004	.076	18±42
497	25	3.80	1.14	.078	.062	.94	.012	.041	.008	.005	.008	.036	.001	.048	54±42
561	28	3.68	1.35	.058	.043	1.01	.010	.054	.004	.018	.013	.012	.001	.030	66±42
587	19	3.73	1.15	.096	.044	.95	.027	.100	.011	.022	.014	.064	.001	.070	61±42
Av. E. A. Plant	70.45	3.79	1.23	.075	.058	.89	.016	.079	.006	.015	.019	.061	.002	.110	71±42

Table 64 Bend Test Results For WEU 100 Moulds.

Mould No.	Position	No. of Specimens	Max. Load kN	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A/B	A	B
404	BO	1	2.8	7.8	11.4	3.1	2.5	0.8
	BM	3	2.9	12.3	12.1	4.5	3.2	0.8
	BI	1	1.9	5.3	6.3	3.1	1.3	0.4
	TO	1	1.6	12.0	23.1	3.1	1.3	0.4
	TM	3	1.9	6.2	7.3	3.3	1.5	0.5
	TI	1	2.1	2.8	3.1	1.8	1.7	0.9
	VO	1	2.2	6.6	6.5	2.8	1.7	0.6
	VM	1	1.4	3.2	2.9	2.0	0.5	0.3
	VI	1	1.8	1.2	6.7	2.0	1.0	0.5
	V(Av)	3	1.8	3.7	5.4	2.3	1.1	0.5
489	VO	1	2.2	7.0	7.1	3.2	1.0	0.3
	VM	1	2.3	8.3	8.8	3.5	1.4	0.4
	VI	1	2.8	3.7	3.9	1.6	0.7	0.4
	V(Av)	3	2.4	6.3	6.6	2.8	1.0	0.4
497	VM	3	1.9	6.7	5.8	2.6	1.4	0.6
561	VM	3	2.1	6.8	7.6	2.4	1.6	0.7
587	VM	3	2.4	10.4	8.3	3.6	1.5	0.4
Good performance Pearlitic mould iron.			1.3-1.6	1.5-3.	2.0-3.5	1.0-1.5		

Table 65 Bend Test Parameters On Horngate Samples From 610 WB Moulds.

Mould No.	Life	Reason for Failure	Max. Load, kN	a/b
2952	97	Crazed	1.84	10.8
2954	80	"	1.73	9.4
2956	89	"	1.55	13.1
2957	109	"	1.35	5.6
2964	88	"	1.99	14.0
2966	84	"	1.55	8.6
2969	97	"	2.08	27.5
2970	82	"	1.79	6.8
2974	96	"	1.73	8.3

Table 66 Bend Test Parameters On Large Lug Samples From Renishaw 610 WB Moulds

Mould No.	Life	Reason for Failure	Max. Load, kN	a/b
3348	111	Crazed	1.52	4.9
3349	88	"	1.64	9.2
3350	80	Cracked face	1.78	8.2
3352	71	Crazed	1.10	2.3
3354	75	"	1.62	5.4
3355	97	"	0.91	1.5
3358	63	"	1.44	5.3
3359	74	"	1.43	3.3
3362	111	"	1.15	3.7

Table 67 Bend Test Parameters On Large Lug Samples Produced At Distington

Mould No.	Life	Reason for Failure	Max. Load,kN	a/b
6506	85	Crazed	1.21	3.0
6508	86	"	1.38	5.8
6516	101	"	1.30	3.4
6522	79	"	1.36	2.8
6693	82	"	1.45	6.2
6700	73	"	1.05	2.5
6704	62	"	1.27	3.3
6728	40	Cracked	1.36	2.8
6730	67	Crazed	1.02	3.7
6744	92	"	1.13	3.0
6766	68	"	1.40	4.8
6767	105	"	1.60	5.3
6785	80	"	1.40	4.7
6787	79	"	1.55	6.0
6791	77	"	1.14	4.8
6792	92	"	1.55	5.2
6799	68	"	1.13	4.1
6809	104	"	1.71	8.1
6808	82	"	1.07	3.8
6813	122	"	1.61	10.7
6828	40	"	1.50	7.9

Table 68 Chemical Analyses And Bend Test Results From Lug Samples On Distington 610WB Moulds

Mould No.	Life	Cast Temp	C	Si	Mn	S	P	Ti	Max. Load	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A	B	A/B
6506	85	1300	3.84	1.45	0.50	0.023	0.042	0.018	1.22	3.0	3.6	0.2	0.1	1.6
6508	86	-	3.92	1.38	0.60	0.031	0.054	0.019	1.38	5.8	6.1	0.2	0.1	2.7
6516	101	1270	3.89	1.21	0.58	0.035	0.046	0.016	1.30	3.4	4.2	0.2	0.1	2.0
6522	79	1280	3.82	1.59	0.64	0.010	0.050	0.026	1.36	2.8	5.1	0.2	0.1	2.1
6693	82	1290	3.87	1.37	0.77	0.033	0.048	0.017	1.45	6.2	6.3	0.2	0.1	2.7
6700	73	1260	4.00	1.19	0.61	0.021	0.042	0.023	1.05	2.5	3.2	0.2	0.1	2.0
6704	62	1280	4.00	1.73	0.57	0.030	0.053	0.043	1.27	3.3	3.5	0.2	0.1	2.0
6728	40	1280	3.81	1.69	0.54	0.044	0.040	0.018	1.36	2.8	5.13	0.2	0.1	2.1
6730	67	1270	4.00	1.48	0.53	0.012	0.044	0.043	1.02	3.7	3.7	0.2	0.1	2.2
6744	92	1290	3.96	1.28	0.54	0.038	0.047	0.030	1.13	3.0	3.1	0.1	0.1	1.4
6760	68	1250	4.00	1.37	0.59	0.019	0.047	0.028	1.40	4.8	5.2	0.2	0.1	2.9
6767	105	1270	3.76	1.23	0.51	0.039	0.046	0.026	1.60	5.3	6.5	0.2	0.1	2.5
6785	80	1250	3.92	1.17	0.52	0.028	0.045	0.022	1.40	4.7	5.1	0.2	0.1	2.7
6787	79	-	3.77	1.29	0.53	0.037	0.042	0.023	1.55	6.0	6.1	0.3	0.1	2.8
6791	77	1290	4.04	1.68	0.69	0.044	0.054	0.041	1.14	4.8	5.0	0.2	0.1	1.8
6792	92	1280	3.92	1.39	0.51	0.039	0.051	0.032	1.55	5.2	7.0	0.2	0.1	2.7
6799	68	1280	4.10	1.63	0.54	0.034	0.045	0.037	1.13	4.1	5.8	0.2	0.1	2.0
6809	104	1260	3.80	1.15	0.47	0.038	0.046	0.019	1.71	8.1	11.3	0.3	0.1	3.1
6808	82	1260	3.94	1.50	0.59	0.029	0.046	0.047	1.07	3.8	6.0	0.2	0.1	2.3
6813	122	1250	3.72	1.14	0.50	0.38	0.039	0.015	1.61	10.7	16.7	0.3	0.1	3.5
6828	40	1280	3.81	1.69	0.54	0.044	0.040	0.018	1.50	7.9	12.0	0.3	0.1	3.2

Table 69 Correlation Matrix Of Variables And Life For Distington Lug Samples

Life	C	Si	Mn	S	P	Ti	Max L	a/b	$A\frac{1}{2}/B\frac{1}{2}$	A	B	A/B
-.048												
-.749*	.334											
-.182	.306	.270										
.067	-.264	.028	-.150									
.110	.450*	.234	.457*	-.077								
-.212	.671*	.474	.098	-.214	.428							
.345	-.748*	-.361	-.289	.409	-.109	-.593*						
.381	-.513*	-.323	-.198	.431	-.216	-.388	.719*					
.319	-.579*	-.219	-.294	.413	-.360	-.395	.670*	.930*				
.006	-.649*	-.113	-.253	.269	-.211	-.524*	.819*	.722*	.717*			
-.251	.015	.406	-.004	-.039	-.162	.075	-.240	-.419	-.384	-.048		
.168	-.497*	-.323	-.163	.209	-.243	-.421	.760*	.869*	.817*	.818*	-.567*	

* - Significant Relationship At 5% Confidence Level If Coefficient $>|0.433|$

1% " " " " $>|0.549|$

0.1% " " " " $>|0.665|$

Table 70 Chemical Analyses Of The Foundry Test Blocks

Foundry	Code	C	Si	Mn	S	P	Ti	Sn	V	Cr	Ni	Mo	Cu
Dowlais	1	3.51	1.68	0.53	0.060	0.08							
	2	3.43	1.58	0.48	0.098	0.042							
	3	3.57	1.58	0.52	0.091	0.058							
Landore	1	3.40	1.69	0.51	0.065	0.032		0.006	0.010	0.011	0.004	0.001	0.032
	2		1.10	0.52	0.065	0.044		0.009	0.005	0.028	0.026	0.001	0.050
Distington	1	3.85	1.33	0.54	0.041	0.052	0.017						
	2	3.99	1.77	0.61	0.029	0.056	0.015						
	3	3.96	1.35	0.75	0.032	0.084	0.032						
	4	3.88	1.18	0.50	0.015	0.044	0.028						
	5	3.99	1.77	0.61	0.029	0.056	0.015						
	6	3.82	1.58	0.66	0.030	0.075	0.031						
	7	3.96	1.57	0.60	0.032	0.047	0.025						
	8	3.82	1.54	0.63	0.044	0.052	0.021						

Table 71 Mechanical Properties Of The Foundry Test Blocks

Foundry	Code	Tensile Strength N/mm^2	Max L. kN	a/b	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	A	B	A/B
Dowlais	1	86	1.6	2.6	3.1	1.2	0.8	1.5
	2	74	1.4	2.5	2.9	1.0	0.8	1.3
	3	68	1.5	2.8	3.6	1.2	0.8	1.5
Landore	1		1.6	3.6	4.2	1.8	1.0	2.1
Distington	1		1.8	2.6	2.7	1.4	1.1	1.4
	2		1.2	2.8	3.3	1.1	0.9	1.4
	3		1.3	2.6	3.0	0.8	0.8	1.0
	4		1.3	2.3	2.8	1.2	0.9	1.3
	5		1.1	1.3	1.5	0.8	1.0	0.9
	6		1.4	2.4	2.9	1.3	1.1	1.4
	7		1.3	2.1	2.7	1.2	1.1	1.2
	8		1.3	2.3	3.1	1.2	0.9	1.3
Fullwood	1	71	1.4	3.4	4.0	1.8	1.1	1.6
	2	71	1.1	1.5	1.9	1.0	1.2	0.7
Craigneuk	1	98	1.6	5.1	5.2	1.3	0.5	2.4
	2	97	1.7	5.7	12.2	1.4	0.6	2.5

Table 72 The Effect Of Casting Temperature On The Performance Of 610WB Moulds

Cast. Temp. Range °C	No. Of Moulds	Life			
		Mean	Min.	Max.	S. D.
1190-1199	1	65			
1200-1209	5	69.60	54	80	9.86
1210-1219	26	74.92	30	99	17.40
1220-1229	36	82.19	38	132	15.85
1230-1239	61	82.23	37	127	17.61
1240-1249	136	86.03	27	140	17.15
1250-1259	167	83.77	19	144	20.48
1260-1269	244	87.48	8	157	19.02
1270-1279	300	89.17	22	141	18.23
1280-1289	253	87.04	16	208	21.56
1290-1299	178	89.83	13	130	19.80
1300-1309	124	85.67	16	128	20.20
1310-1319	40	92.95	11	140	22.35
1320-1329	23	87.74	42	121	19.46
1330-1339	11	97.27	74	117	14.50
1340-1349	7	83.29	56	95	13.74
1350-1359	3	87.33	79	95	8.02
1360-1369	1	100			

Table 73 Effect Of Casting Temperature On Failure Mode For 610WB Moulds

Failure % \ Temp. Range	1210-1219	1220-1229	1230-1239	1240-1249	1250-1259	1260-1269	1270-1279	1280-1289	1290-1299	1300-1309	1310-1319	1320-1329
Crazed	88	80.5	85	93	86	86.5	89	85	83	84	80	96
Craze/Crack	4	11	10	3	5	5.5	5	8	8	4	10	-
Crack-Corners	4	5.5	5	1	2	5	5	3.5	6	8	5	4
Crack-Face	-	3	-	1	5	1	1	1	1.5	1.5	2.5	-
Cracked	-	-	-	-	-	1	-	1	-	1	-	-
Mech. Damage	4	-	-	2	2	1	-	1.5	1.5	1.5	2.5	-

Table 74 Production Details According To Failure Mode For 610WB Moulds

	Cracked Corners					Crazed					Cracked & Crazed				
	No.	Mean	Min.	Max.	S.D.	No.	Mean	Min.	Max.	S.D.	No.	Mean	Min.	Max.	S.D.
Cast.Temp.	73	1275	1210	1330	24.27	1400	1268	1190	1360	24.36	95	1271	1210	1420	28.41
C	73	3.84	3.58	4.26	0.120	1400	3.88	3.33	4.32	0.101	95	3.87	3.70	4.15	0.099
Si	73	1.37	1.11	1.86	0.175	1400	1.40	0.90	2.34	0.202	95	1.37	1.03	2.03	0.183
Mn	73	0.59	0.51	0.93	0.085	1400	0.58	0.42	1.10	0.071	95	0.58	0.46	0.85	0.060
S	73	0.041	0.021	0.10	0.012	1400	0.036	0.002	0.11	0.009	95	0.038	0.012	0.087	0.011
P	73	0.048	0.037	0.07	0.006	1400	0.048	0.017	0.11	0.006	95	0.048	0.038	0.067	0.005
Ti	73	0.018	0.007	0.04	0.006	1400	0.022	0.006	0.093	0.008	95	0.021	0.011	0.043	0.007
Life	73	90.04	37	134	18.87	1400	87.32	22	145	17.72	95	96.03	16	208	24.58
	Cracked Face					Mechanical Damage					Cracked				
	No.	Mean	Min.	Max.	S.D.	No.	Mean	Min.	Max.	S.D.	No.	Mean	Min.	Max.	S.D.
Cast.Temp.	24	1267	1220	1310	22.20	19	1268	1210	1310	26.51	6	1277	1260	1300	15.06
C	24	3.91	3.71	4.08	0.098	19	3.89	3.76	4.09	0.090	6	3.92	3.70	4.00	0.114
Si	24	1.47	1.07	2.24	0.268	19	1.38	1.07	1.87	0.248	6	1.61	1.31	1.93	0.255
Mn	24	0.57	0.50	0.72	0.053	19	0.59	0.45	0.92	0.093	6	0.58	0.52	0.66	0.053
S	24	0.038	0.015	0.06	0.009	19	0.038	0.026	0.056	0.007	6	0.042	0.034	0.063	0.011
P	24	0.047	0.038	0.07	0.006	19	0.049	0.035	0.088	0.011	6	0.046	0.044	0.050	0.002
Ti	24	0.023	0.006	0.07	0.012	19	0.019	0.010	0.031	0.006	6	0.028	0.021	0.036	0.005
Life	24	67.08	40	111	17.53	19	30.42	8	77	17.91	6	78.83	56	102	18.66

Table 75 t-Tests On Failure Mode For 610WB Moulds - Crazed V Other Modes

	Cracked Corners			Crack/Craze			Cracked Face			Mech. Damage			Cracked		
	Comb.σ	t	Sig.	Comb.σ	t	Sig.	Comb.σ	t	Sig.	Comb.σ	t	Sig.	Comb.σ	t	Sig.
Cast.Temp	24.36	2.39	*	24.64	1.15	n/s	24.33	0.20	n/s	24.39	0	n/s	24.33	0.90	n/s
C	0.102	3.02	**	0.101	0.19	n/s	0.101	1.40	n/s	0.101	0.73	n/s	0.101	0.99	n/s
Si	0.201	1.33	n/s	0.201	1.69	n/s	0.204	1.57	n/s	0.203	0.53	n/s	0.203	2.46	*
Mn	0.072	1.18	n/s	0.070	0.83	n/s	0.079	0.55	n/s	0.071	0.25	n/s	0.071	0.17	n/s
S	0.009	4.54	***	0.009	2.07	*	0.009	0.86	n/s	0.009	0.63	n/s	0.009	1.57	n/s
P	0.006	0.14	n/s	0.006	0.32	n/s	0.006	0.90	n/s	0.006	0.65	n/s	0.006	0.66	n/s
Ti	0.008	4.68	***	0.008	0.94	n/s	0.008	0.77	n/s	0.008	1.39	n/s	0.008	1.84	n/s
Life	17.78	1.27	n/s	18.23	4.51	***	17.72	5.55	***	17.72	13.9	***	17.72	1.17	n/s

n/s - Not Significant

* - Significant At 5% Probability Level

** - " " 1% " "

*** - " " 0.1% " "

Table 76 Effect Of Casting Temperature in L26 Moulds.

Temp. Interval	N	% Crazed	% V.C.	% Sticker	Life
1225-1249	1	100	-	-	-
1250-1274	79	63.3	22.8	13.9	79.4
1275-1299	66	77.3	13.6	9.1	74.0
1300-1324	21	66.7	23.8	9.5	87.8

Table 77 Production Details According To Failure Mode For L26 Moulds

	Crazed					Vertical Cracks				
	No.	Mean	Min.	Max.	S.D.	No.	Mean	Min.	Max.	S.D.
Cast. Temp.	116	1278	1240	1320	16.08	32	1275	1250	1320	16.87
C	135	3.88	3.67	4.13	0.098	40	3.87	3.70	4.05	0.095
Si	135	1.37	0.89	1.92	0.168	40	1.40	1.10	1.82	0.174
Mn	135	0.62	0.45	0.97	0.104	40	0.65	0.52	0.96	0.117
S	135	0.036	0.008	0.087	0.012	40	0.034	0.007	0.077	0.011
P	135	0.042	0.024	0.072	0.006	40	0.042	0.029	0.054	0.006
Life	136	77.70	2	151	26.90	40	82.83	15	148	24.60
	Sticker					All Data				
	No.	Mean	Min.	Max.	S.D.	No.	Mean	Min.	Max.	S.D.
Cast. Temp.	19	1275	1250	1302	13.78	167	1277	1240	1320	15.95
C	21	3.88	3.73	4.09	0.110	196	3.88	3.67	4.13	0.098
Si	21	1.49	1.11	2.17	0.329	196	1.39	0.89	2.17	0.195
Mn	21	0.66	0.50	0.98	0.128	196	0.63	0.45	0.98	0.110
S	21	0.04	0.01	0.06	0.011	196	0.04	0.01	0.09	0.012
P	21	0.04	0.03	0.06	0.006	196	0.04	0.02	0.07	0.006
Life	21	73.9	38	108	22.68	197	78.3	2.0	151	26.05

Table 78 t-Tests On Failure Modes For L26 Moulds.

	Crazed V V.C.				Crazed V Sticker				V.C. V Sticker			
	D.F,	Comb.σ	't'	sig.	D.F.	Comb.σ	't'	sig.	D.F.	Comb.σ	't'	sig.
Life	174	26.40	1.08	n/s	155	26.39	0.61	n/s	59	23.97	1.37	n/s
T.C.	173	0.097	0.69	n/s	154	0.099	0.00	n/s	59	0.100	0.44	n/s
Mn	173	0.107	1.52	n/s	154	0.107	1.58	n/s	59	0.121	0.32	n/s
S	173	0.012	1.08	n/s	154	0.012	0.19	n/s	59	0.011	0.57	n/s
P	173	0.006	0.60	n/s	154	0.006	1.24	n/s	59	0.006	1.59	n/s
Si	173	0.169	1.05	n/s	154	0.197	2.54	n/s	59	0.238	1.34	n/s
Temp	146	16.25	0.92	n/s	133	15.79	0.77	n/s	49	15.81	0.00	n/s

n/s- Not Significant.

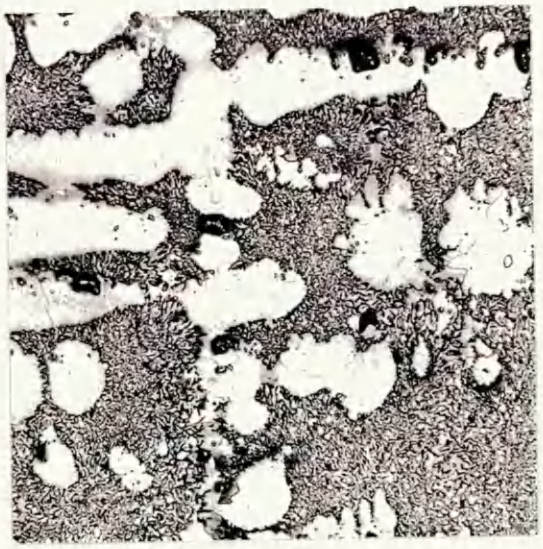
Table 79 Summary Of Bend Test Results.

Group	A+B	A/B	$A^{\frac{1}{2}}/B^{\frac{1}{2}}$	a/b
1) Graphite Coarseness (Table 3-Samples 3-5)	approx. 1.9	0.9-2.8	2.0-8.0	2.5-9.2
2) Undercooled Graphite (Table 3-Samples 1-2)	1.5-2.2	5.8-200	20-100	20-40
3) Variations in Pearlite Content (Table 4)	1.7-2.0	1.0-1.5	2.4-3.9	1.7-2.9
4) Carbides in Flake Graphite (Table 5)				
5) Graphite Morphology(Flake, compacted & Q/F) (Table 6)	2.0-10.4	1.0-2.5	2.4-3.5	1.7-3.5
6) Carbides in Q/F iron (Table 7)	10 - 12	2.5-6.3	3.4-11.5	3.5-10.9
7) Horngates (Table 65)				5.6-27.5
8) Lug Samples (Tables 66-67)				2.3-10.7
9) Phosphorus (Table 9)	1.8-2.2	0.8-4.5	1.6-8.9	1.1-14.3
10) Stripping Practice (Table 9)	1.8-2.3	0.8-6.7	1.6-15.7	1.1-17.5
11) Titanium Heavy Sections (Table 14)	1.7-3.2	0.9-2.0	2.1-6.0	1.8-4.7
12) Titanium Light Sections (Table 13)	1.5-2.0	1.3-100	2.9-100	2.3-40
13) Nitrogen (Table 20)	1.2-2.2	1.0-2.9	2.2-5.7	1.8-6.7
14) Casting Temperature (Table 24)	1.4-1.5	1.0-1.6	2.0-4.1	1.8-3.3
15) Superheat (Table 26)	1.5-2.5	1.7-2.0	2.9-5.5	3.1-4.0
16) Residual Levels (Table22)	1.3-2.7	1.6-2.5	3.0-6.2	2.4-5.7

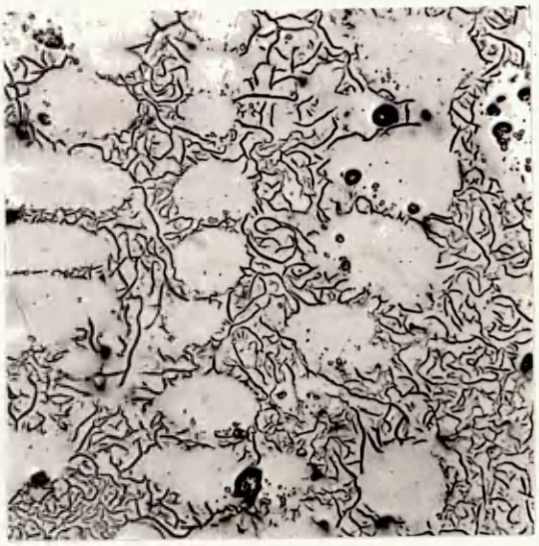
Figure 1 Bend Test System

a) Attachment On Instron

b) Attachment



Sample 1



Sample 2

Sample 3

Sample 4

Sample 5

Figure 2

Graphite Size For Bend Tests

x100

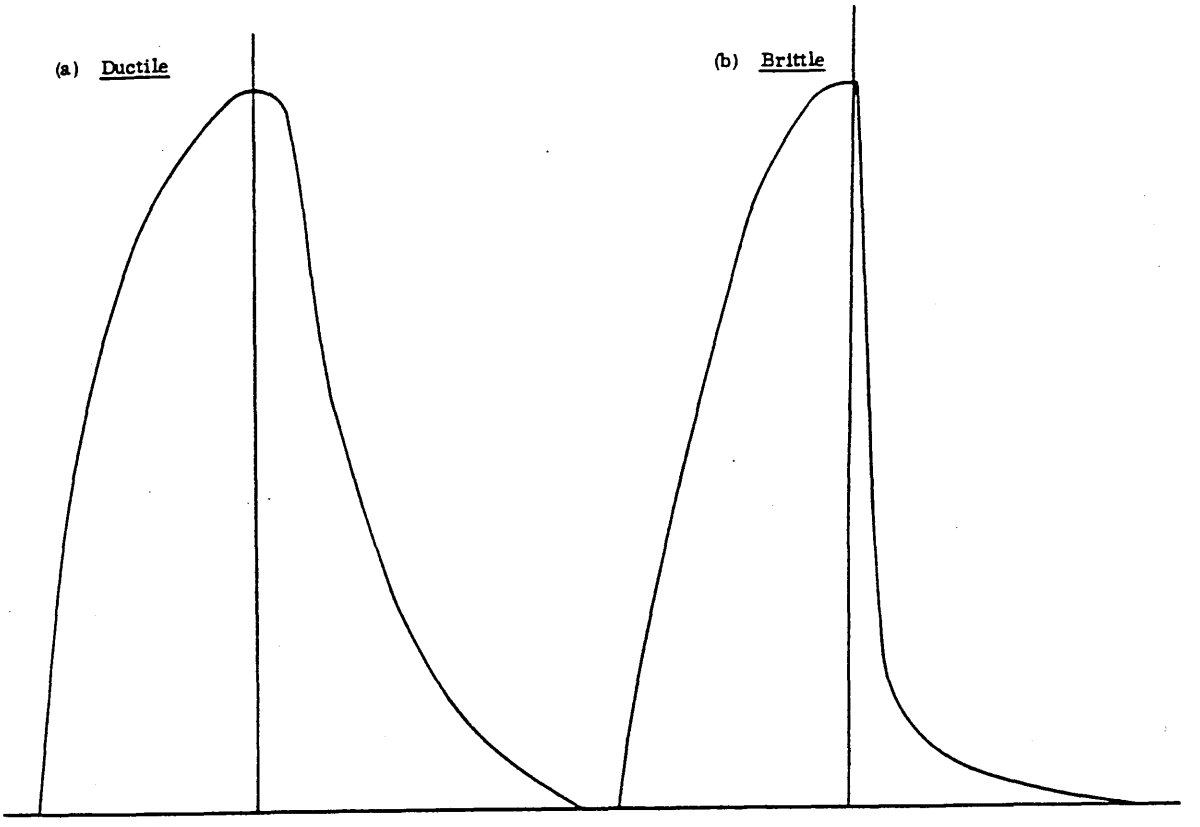


Figure 3 Load-Deflection Curves For Ductile And Brittle Irons

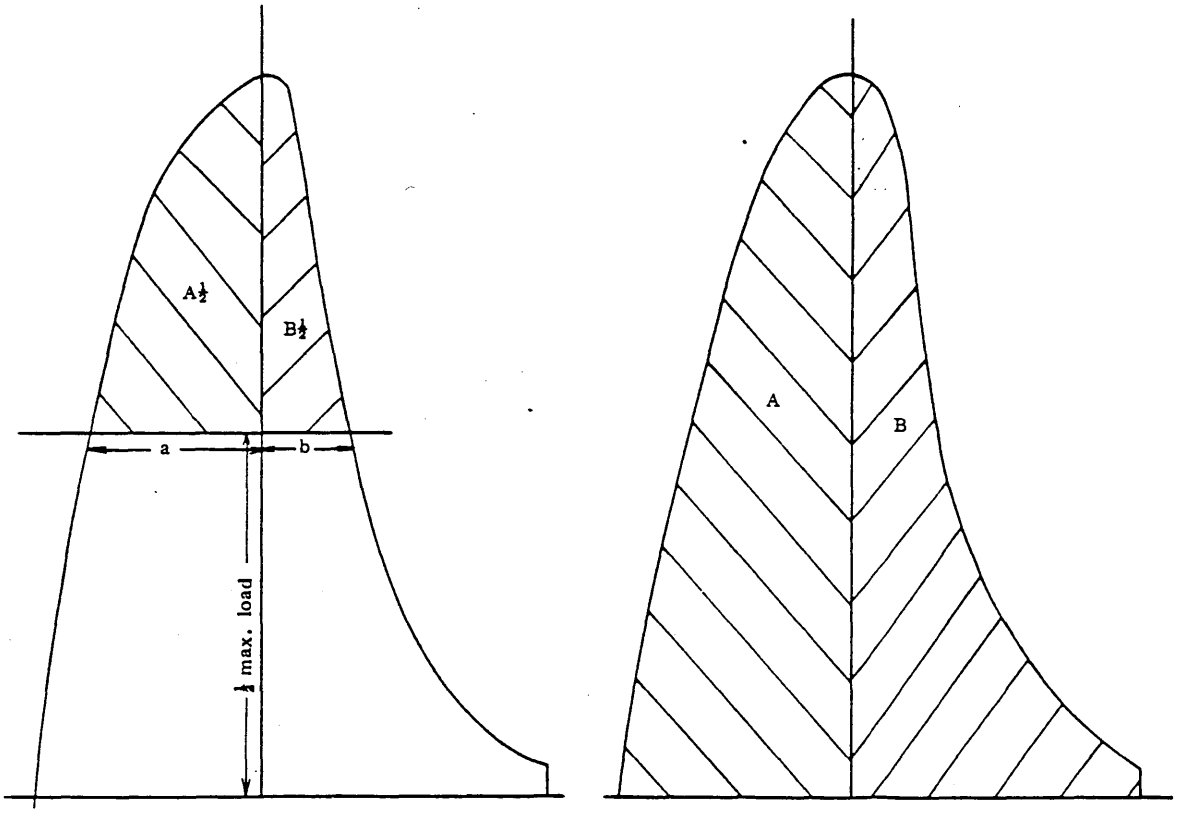


Figure 4 Determination Of Bend Test Parameters

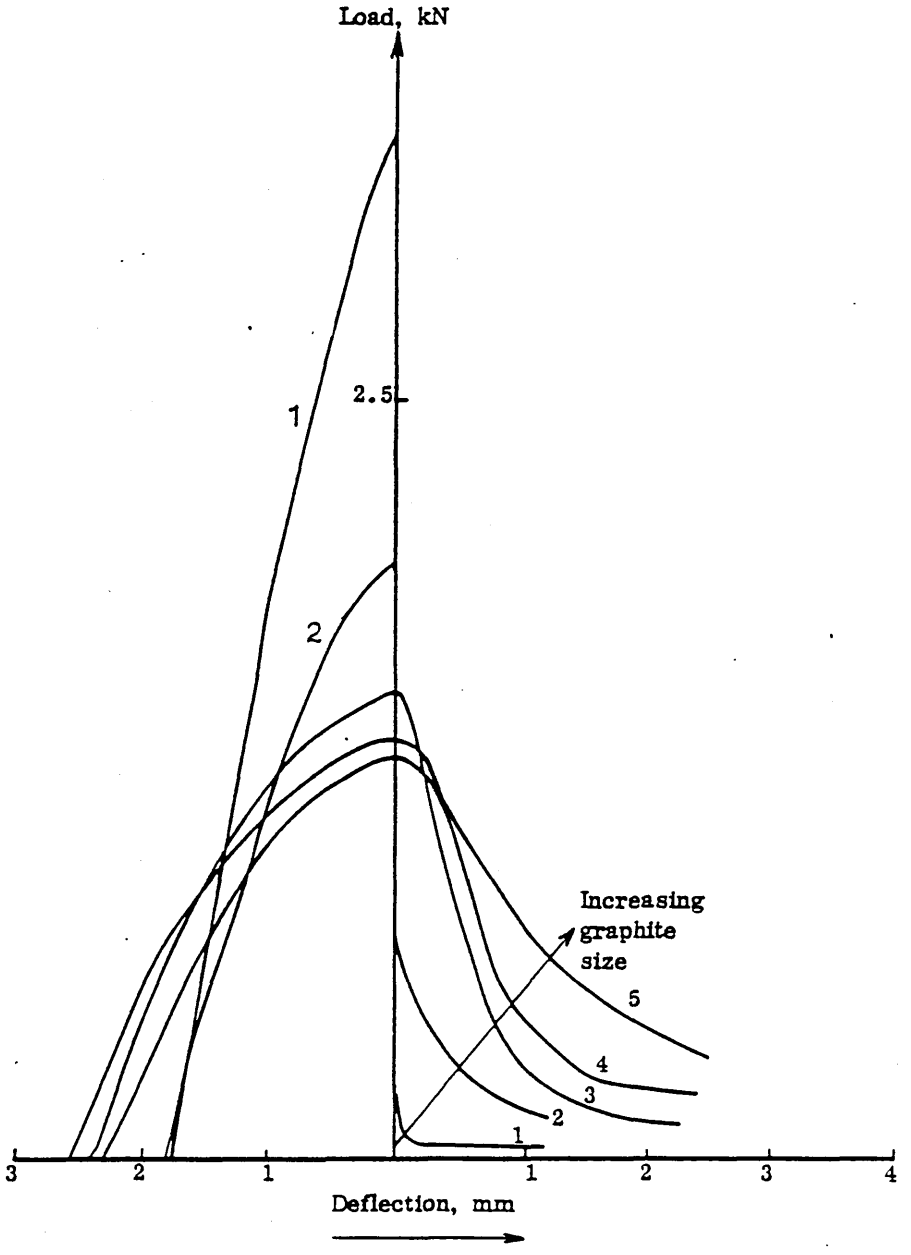


Figure 5 Effect Of Graphite Size On Load-Deflection Curves

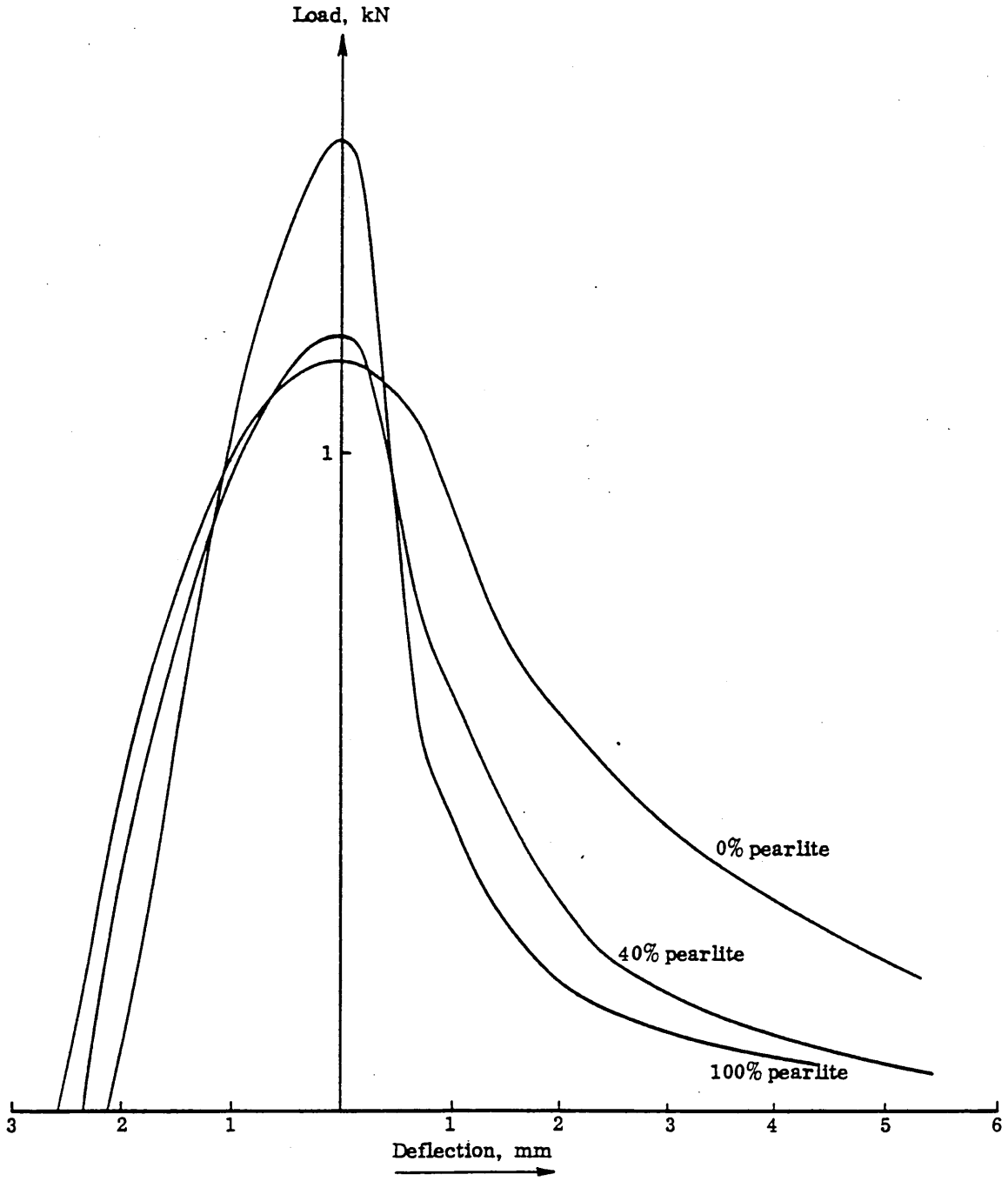


Figure 6 Load-Deflection Curves Showing The Effect Of Pearlite

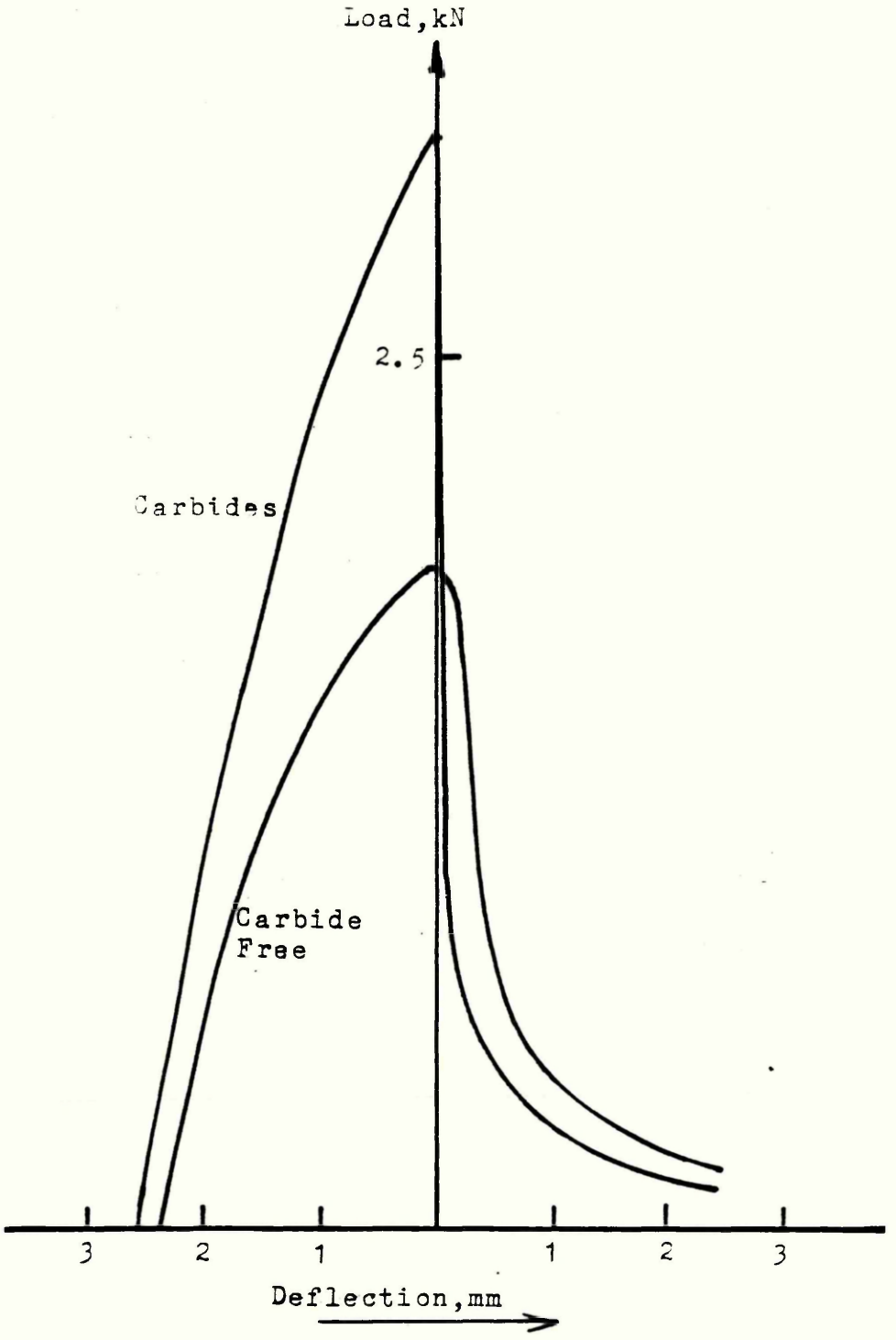


Figure 7 Effect Of Carbides In Flake Grahite Iron

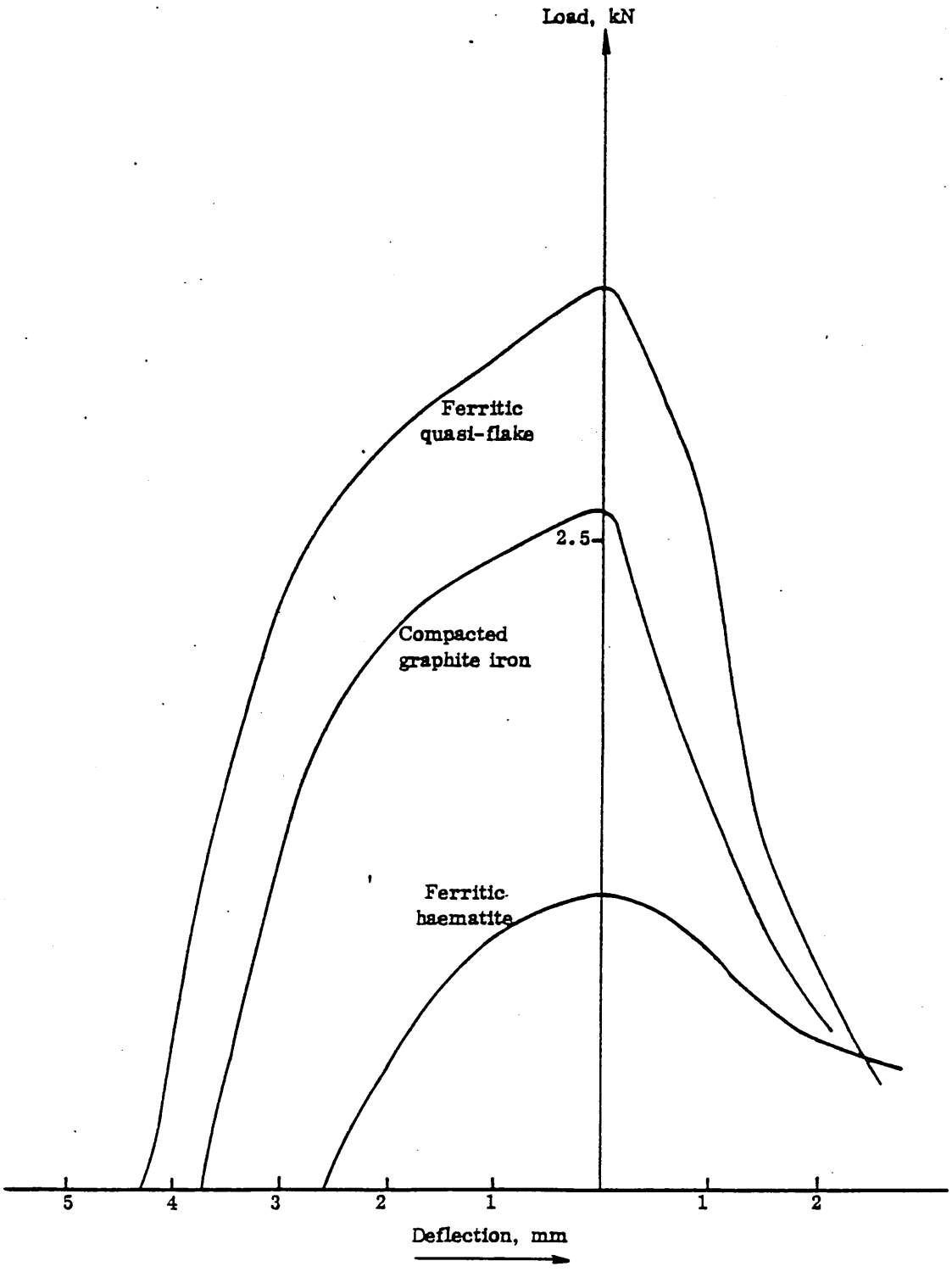


Figure 8 Effect Of Graphite Morphology on Load-Deflection Curves

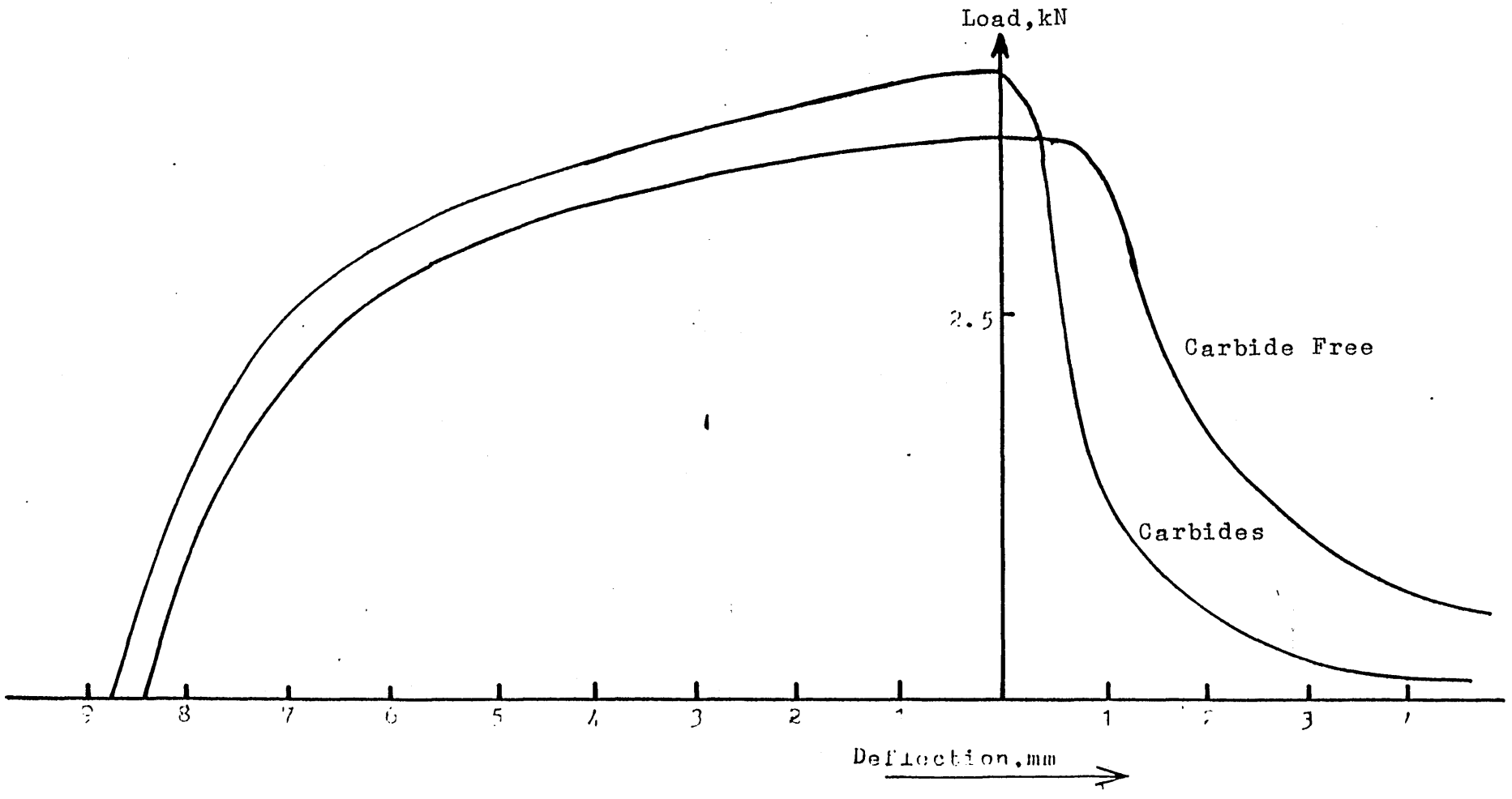


Figure 9 Bend Test Curve On Quasi-Flake Irons With And Without Carbides

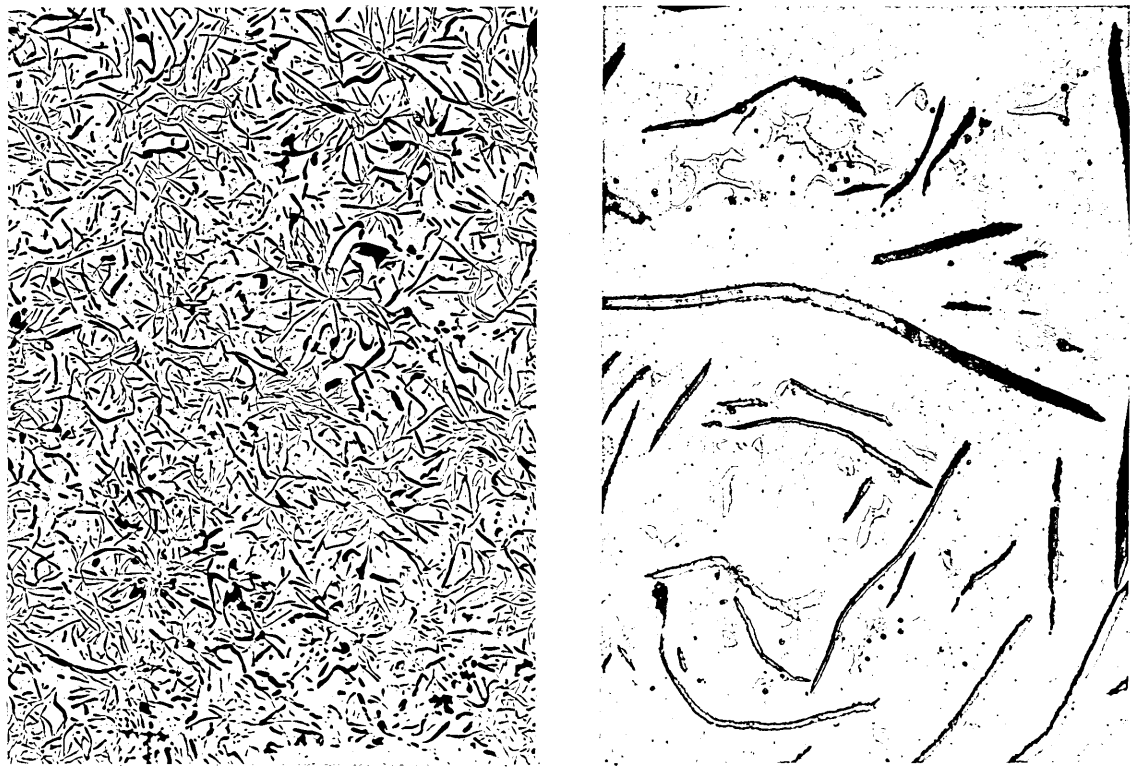
x10 x100

RG202B Slow Cooled

x10 x100

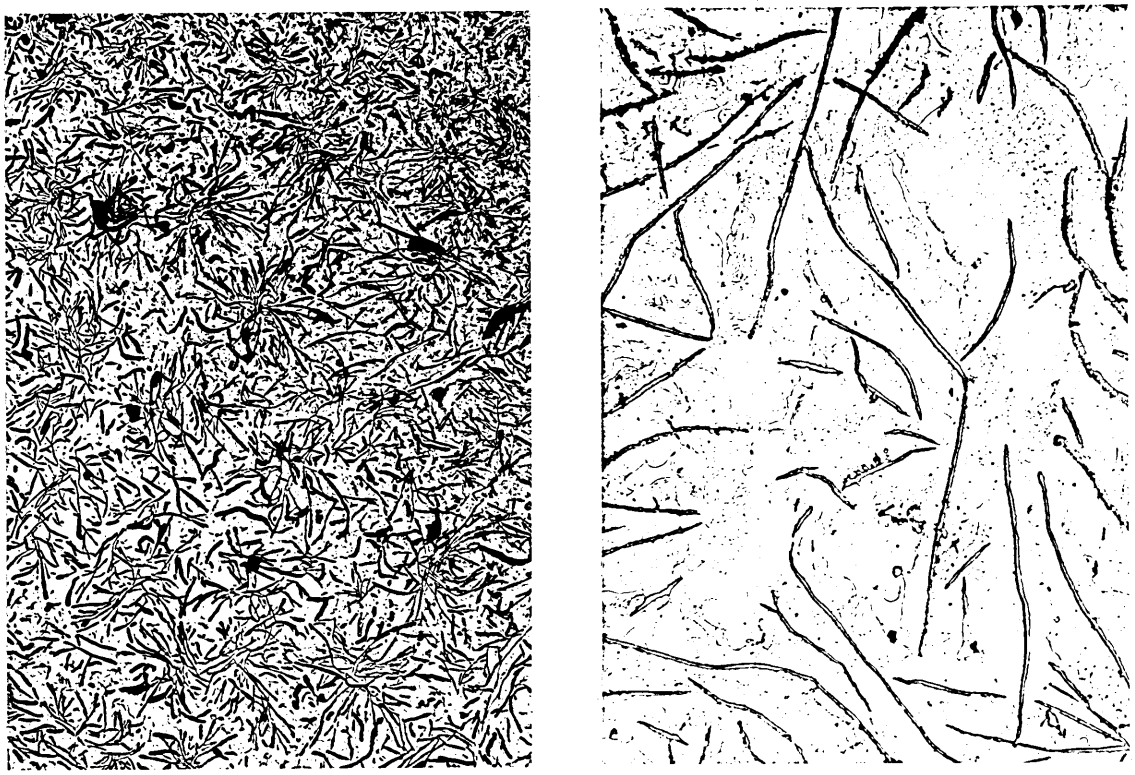
RG202A Fast Cooled

Figure 10 Microstructure Of Low Residual, Low Phosphorus Test Blocks



x10 x100

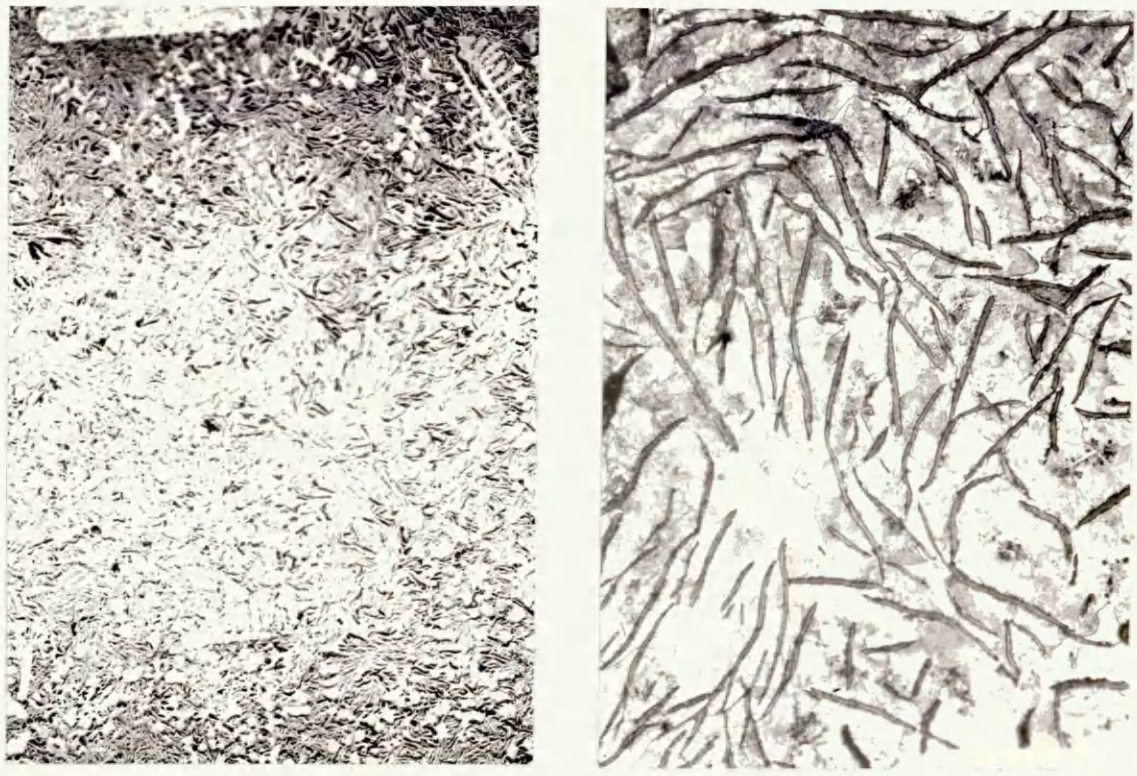
RG203B Slow Cooled



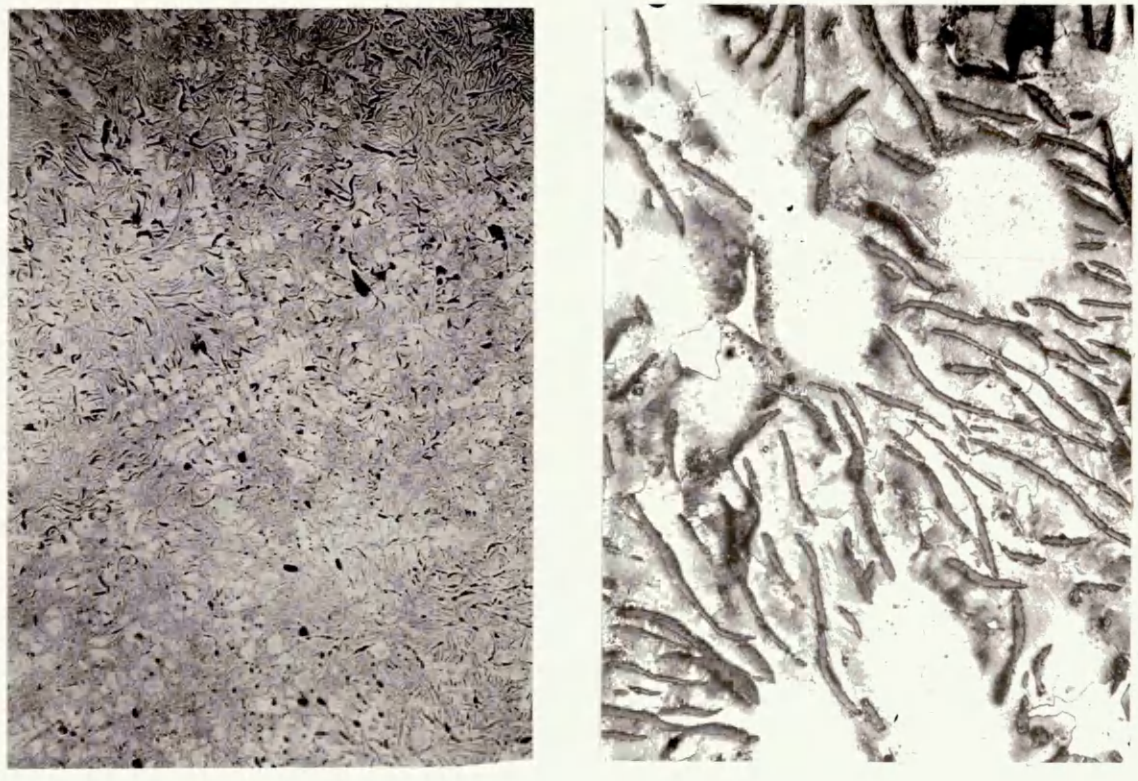
x10 x100

RG203A Fast Cooled

Figure 11 Microstructures Of Low Residual, High Phosphorus Test Blocks



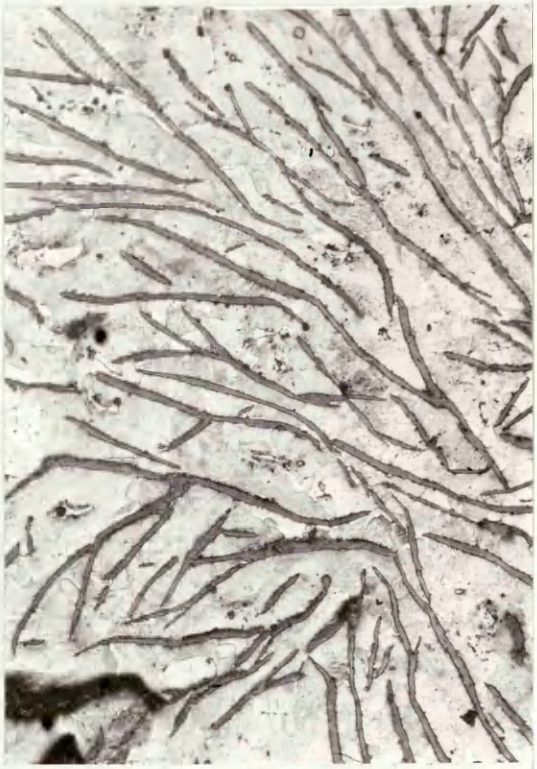
x10 x100
RG7⁄3A Slow Cooled



x10 x100
RG7⁄2A Fast Cooled

Figure 12 Microstructures Of High Residual, Low Phosphorus Test Blocks

x10 x100
RG743B Slow Cooled



x10 x100
RG742B Fast Cooled

Figure 13 Microstructures Of High Residual, High Phosphorus Test Blocks

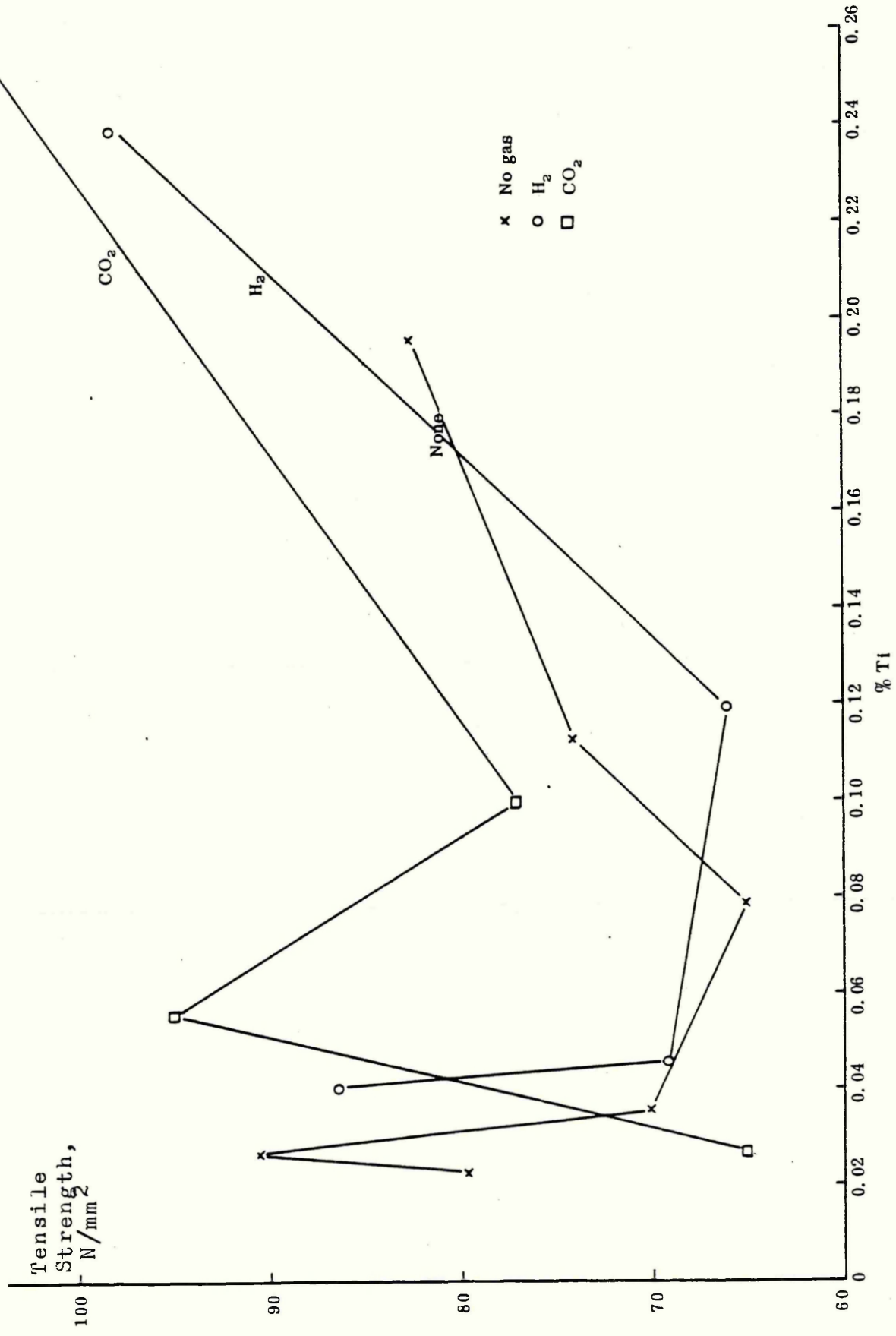


Figure 14 The Effect Of Titanium On Tensile Strength In Heavy Section Test Blocks

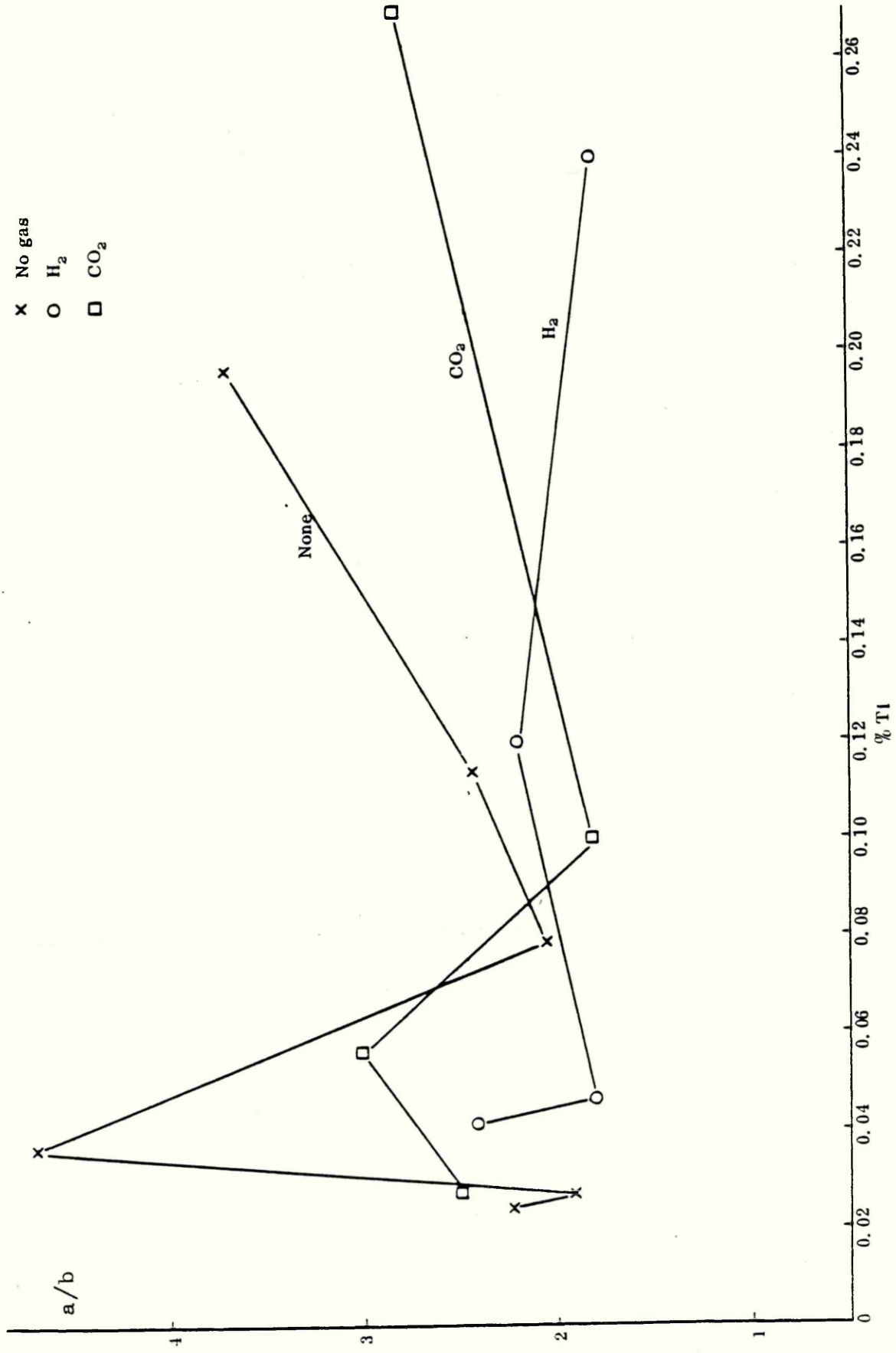


Figure 15 The Effect Of Titanium On The a/b Bend Test Parameter In Heavy Section Test Blocks

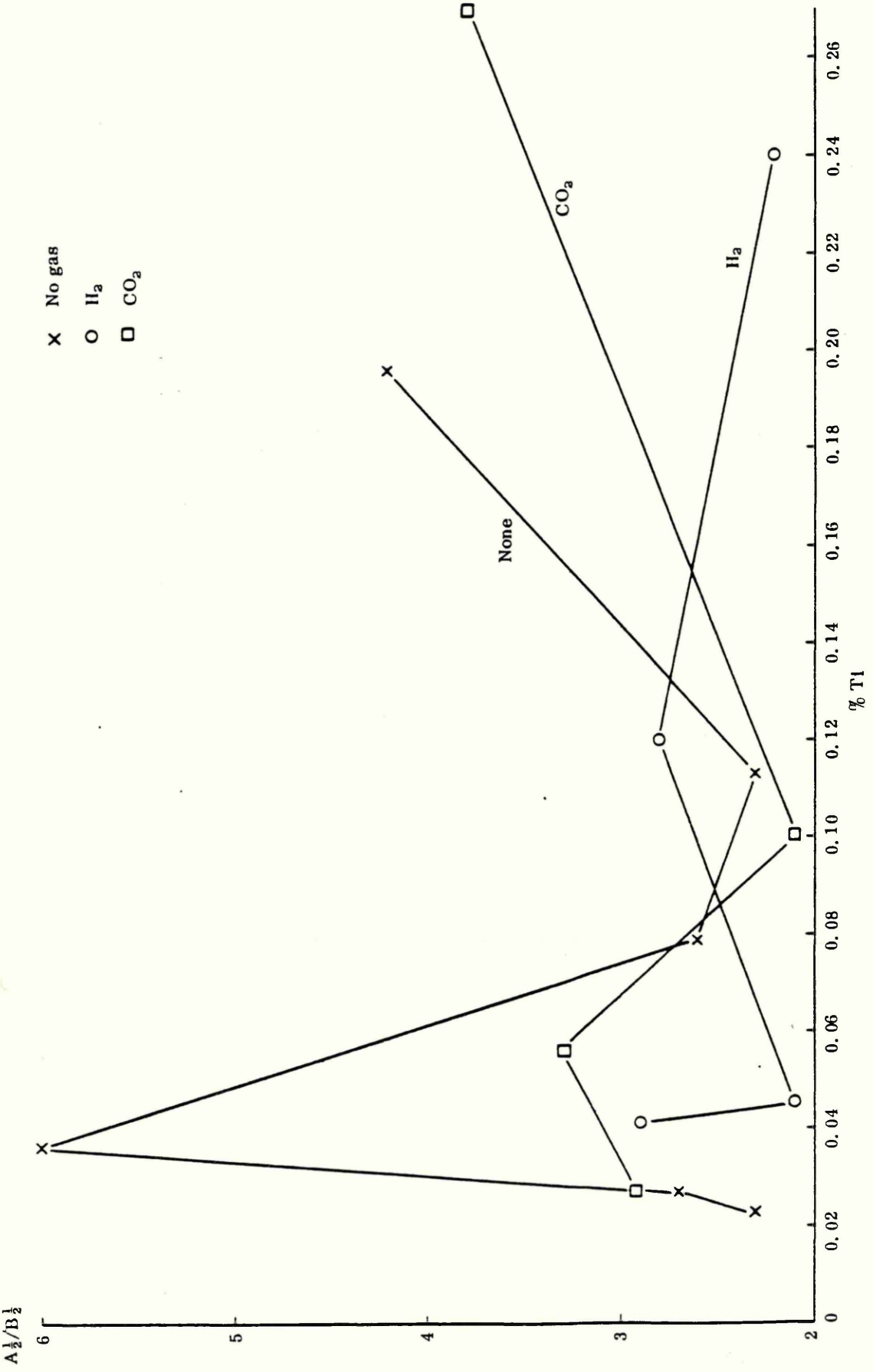


Figure 16 The Effect Of Titanium On The $A\frac{1}{2}/B\frac{1}{2}$ Bend Test Parameter In Heavy Section Test Blocks

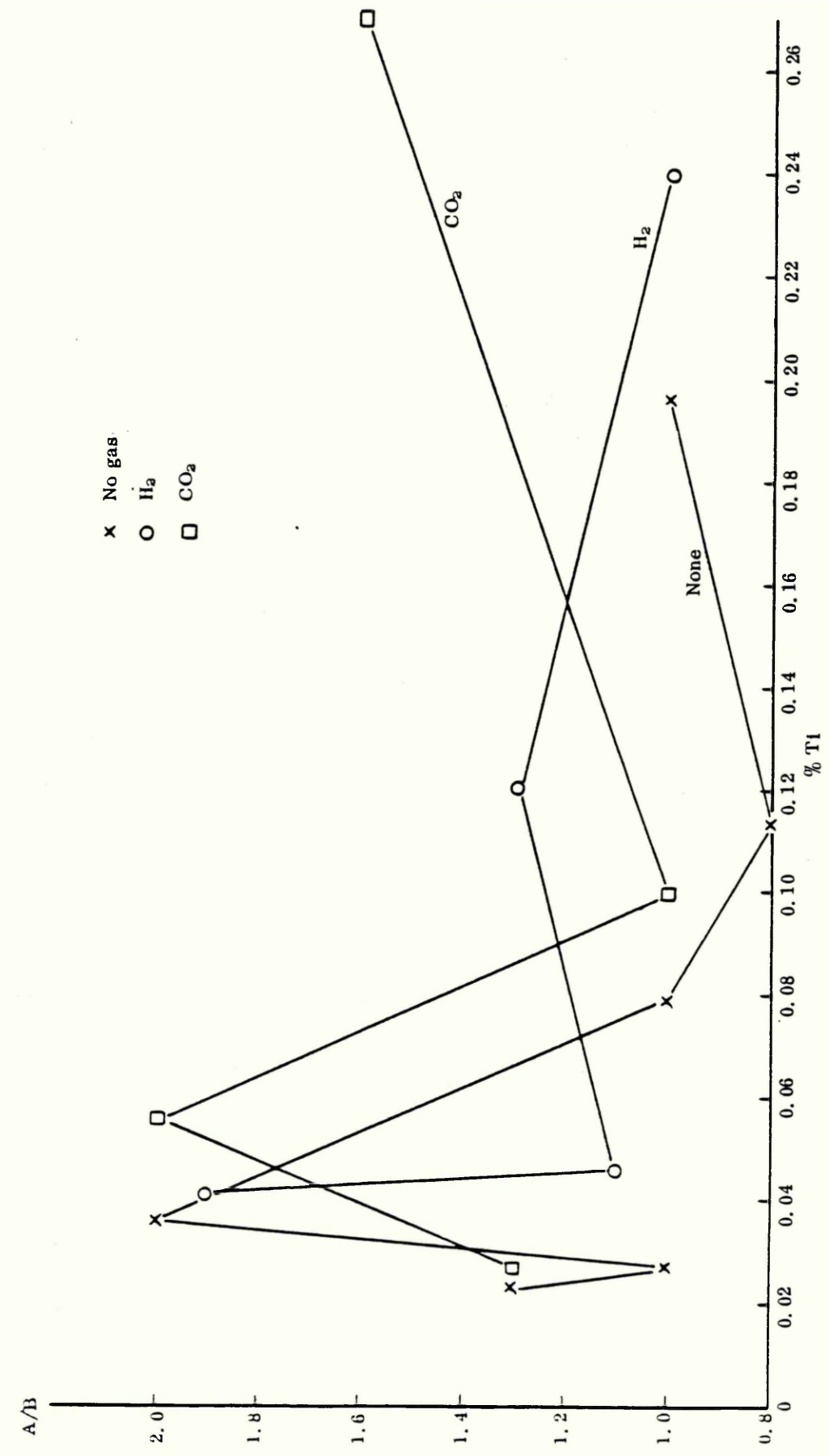


Figure 17 The Effect Of Titanium On The A/B Bend Test Parameter In Heavy Section Test Blocks

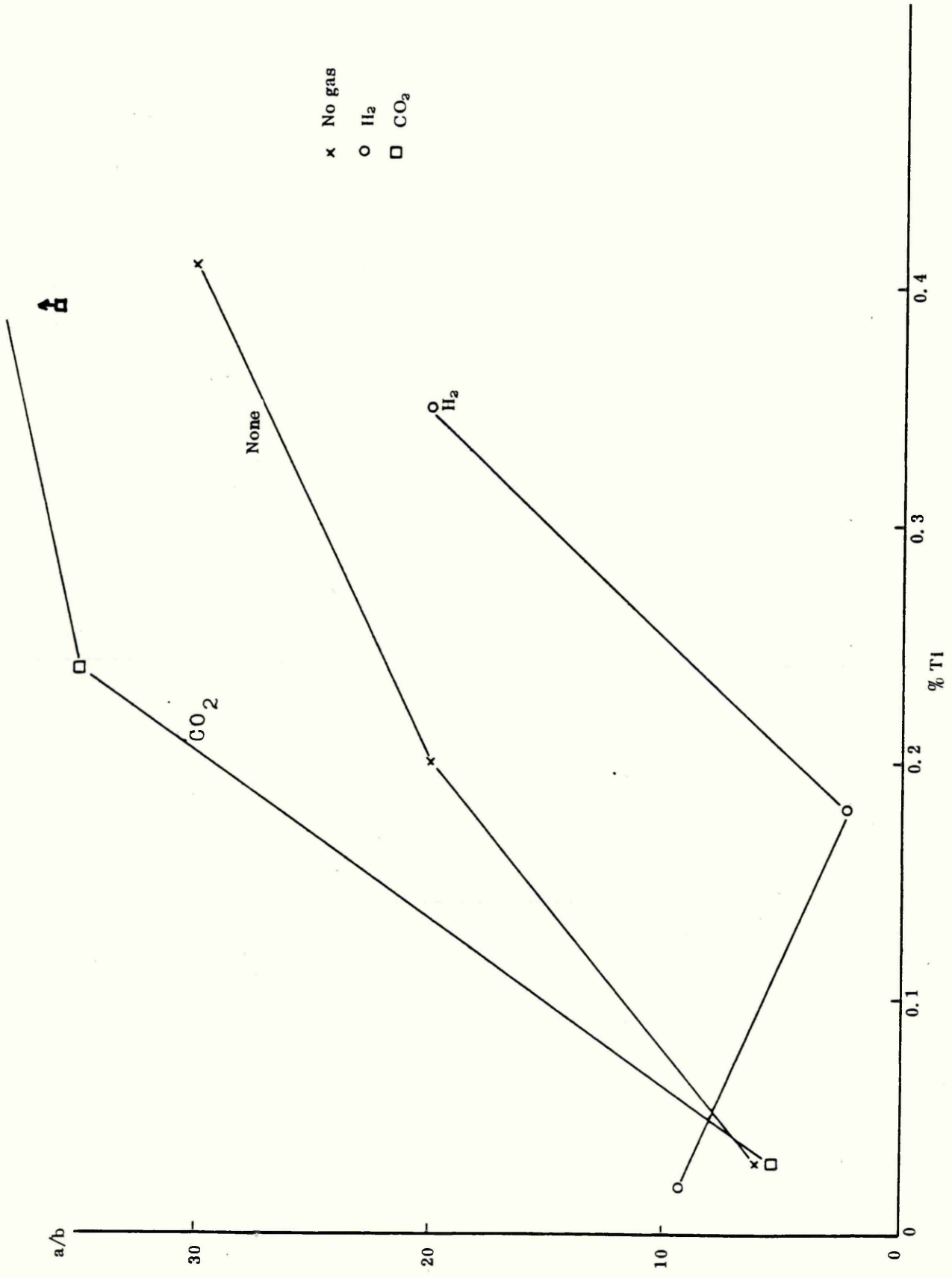
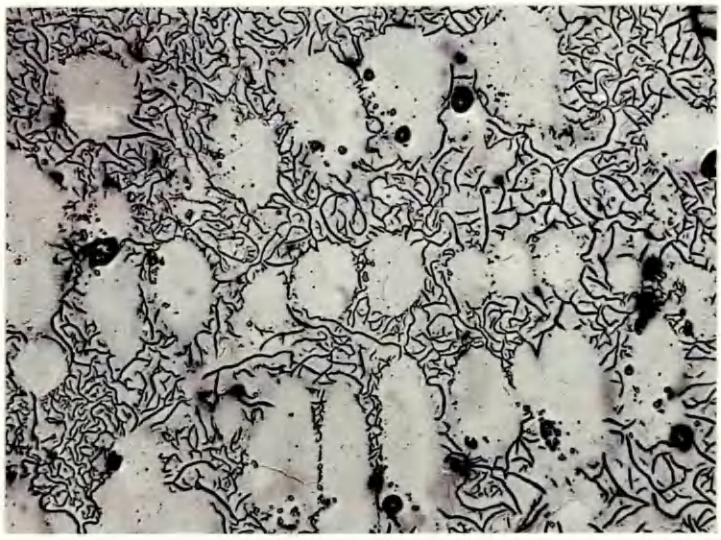


Figure 18 The Effect Of Titanium On The a/b Bend Test Parameter In Light Section Test Blocks

0.02% Ti

0.1% Ti



0.35% Ti

Figure 19(a) Hydrogen Treated Light Section Melts

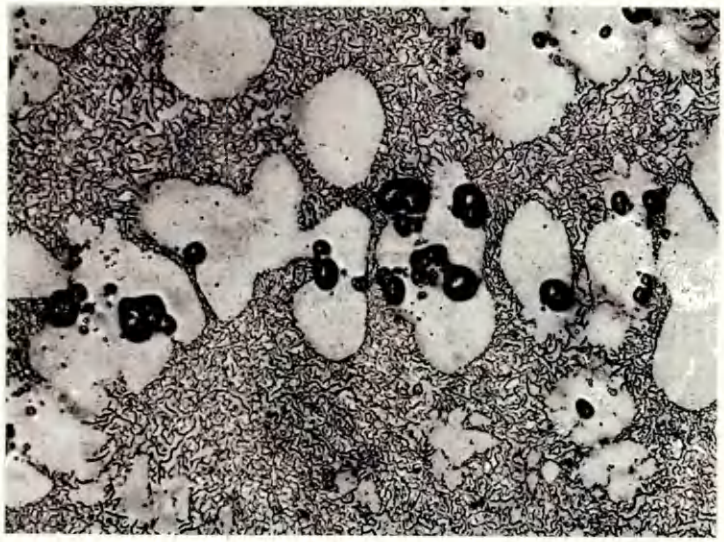
x100

0.03% Ti

Coarse Area Fine Area

0.20% Ti

Coarse Area Fine Area



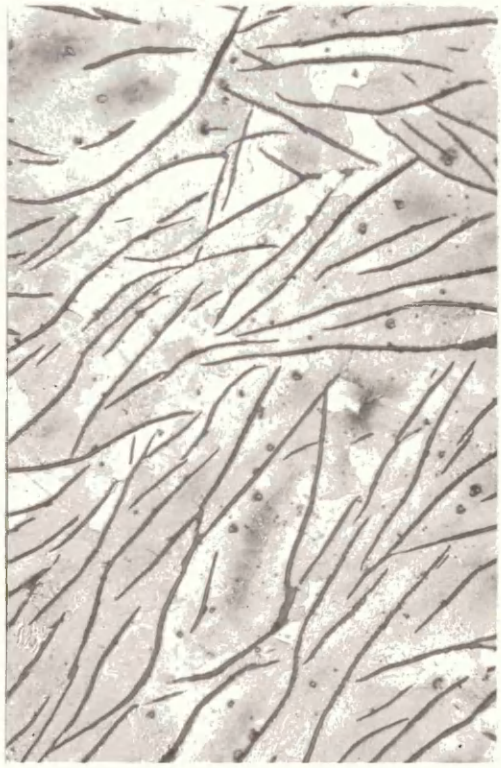
0.4% Ti

Figure 19(b) Untreated Light Section Melts

x100

x10 x100

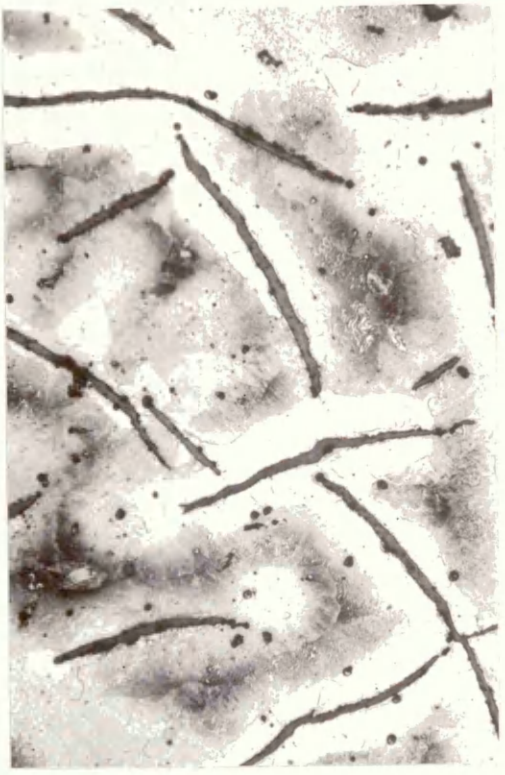
a) RG718 - 0.041% Ti



x10 x100

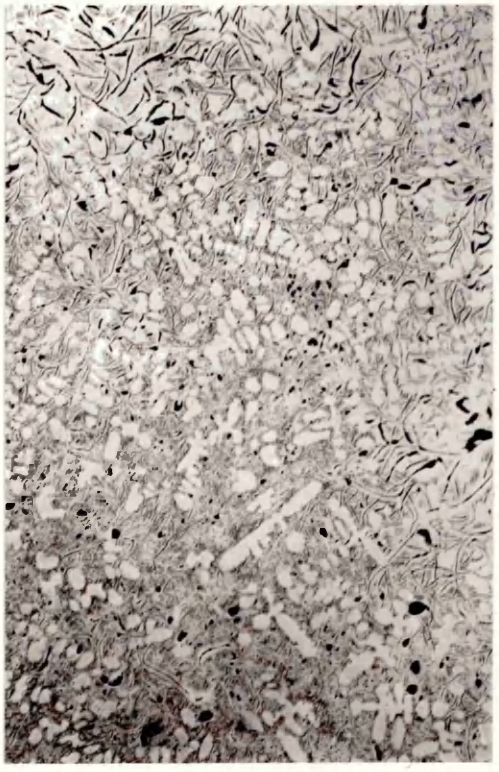
b) RG724 - 0.046% Ti

Figure 20 (Cont'd)



x10 x100

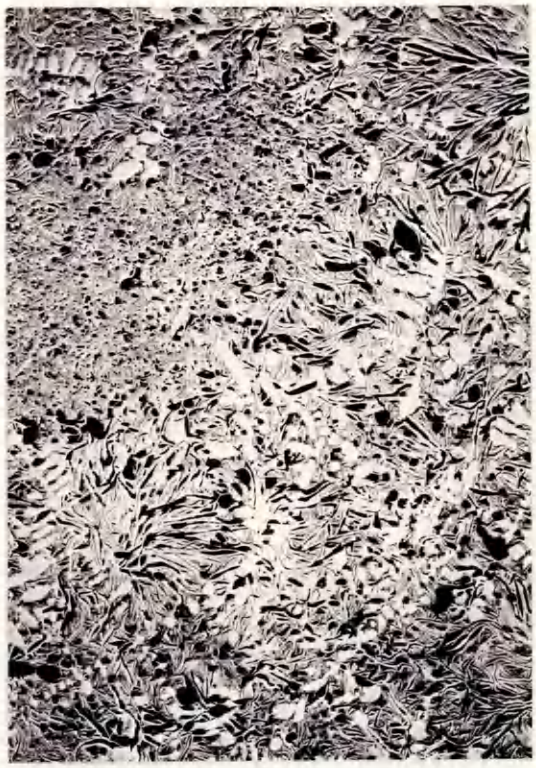
c) RG769 - 0.12% Ti



x10 x100

d) RG754 - 0.24% Ti

Figure 20 Micrographs Of Hydrogen Treated Heavy Section Test Blocks



x10 x100

a) RG300 - 0.023% Ti

x10 x100

b) RG305 - 0.027% Ti

Figure 21 (Cont'd)

x10 x100

c) RG301 - 0.036% Ti

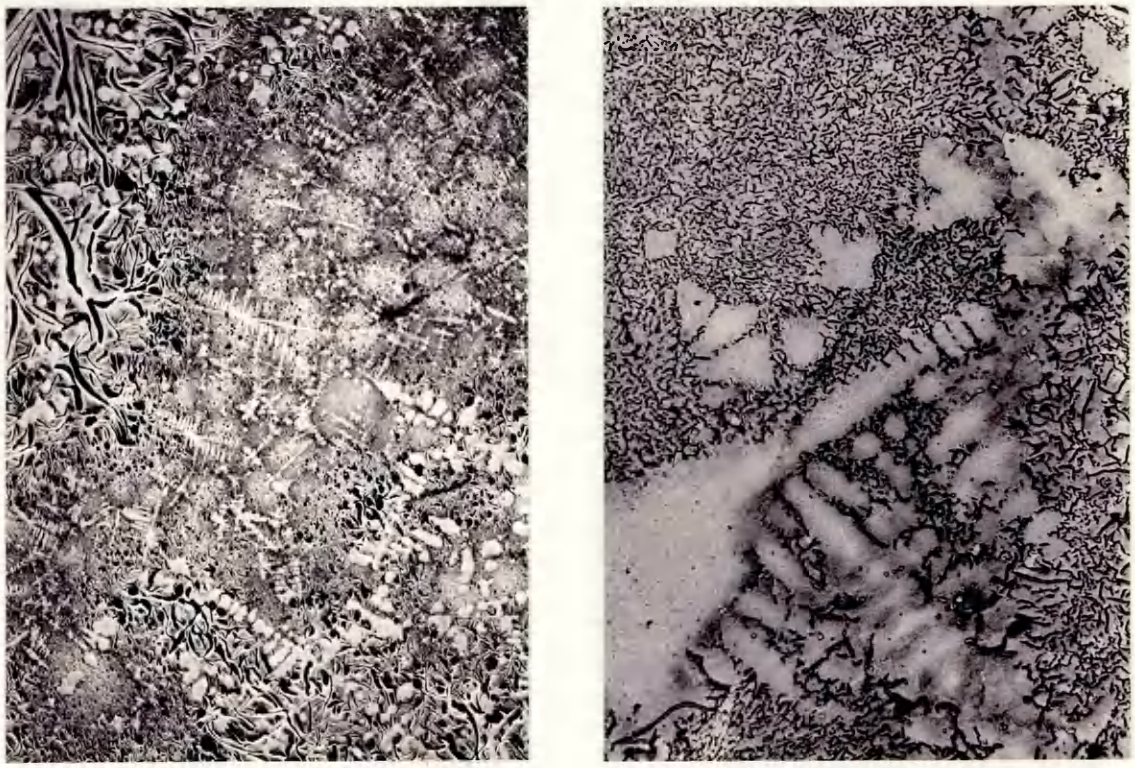
x10 x100

d) RG340 - 0.079% Ti

Figure 21 (Cont'd)

x10 x100

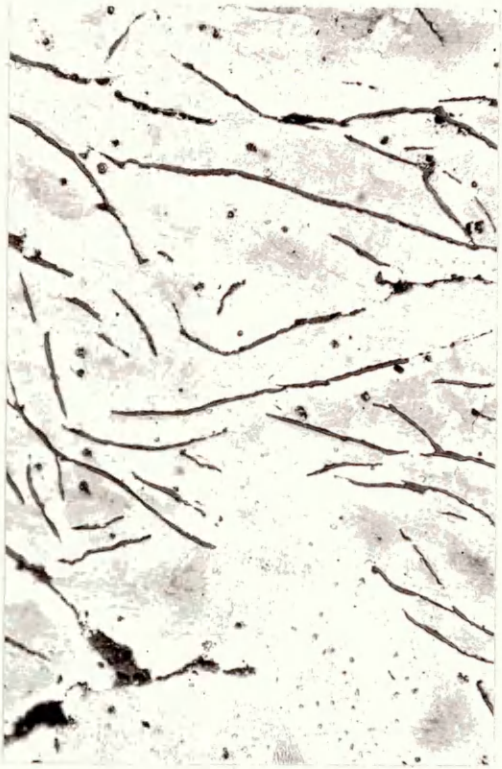
e) RG344 - 0.113% Ti



x10 x100

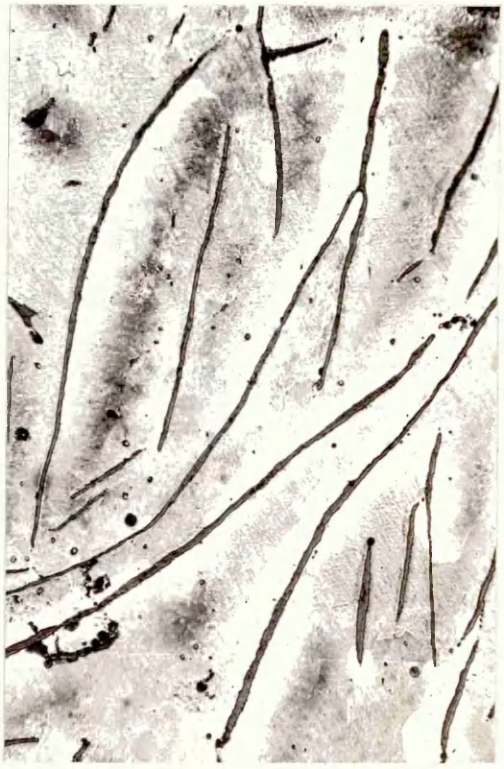
f) RG341 - 0.196% Ti

Figure 21 Micrographs Of Untreated Heavy Section Test Blocks



x10 x100

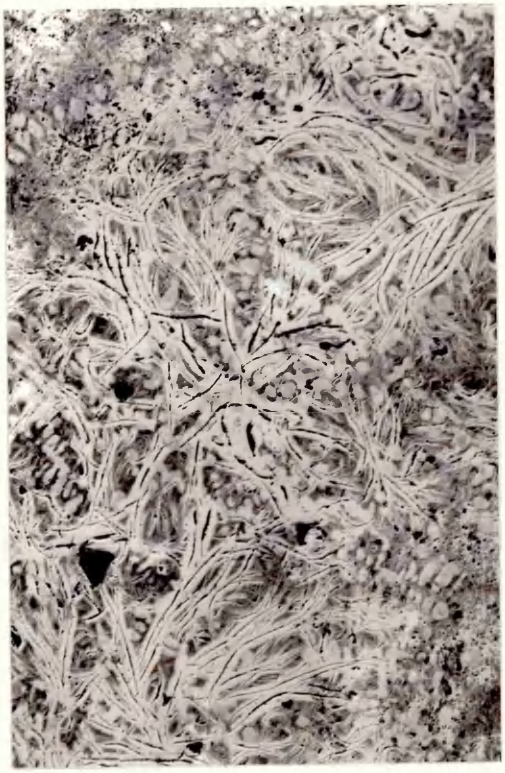
a) RG725 - 0.027% Ti



x10 x100

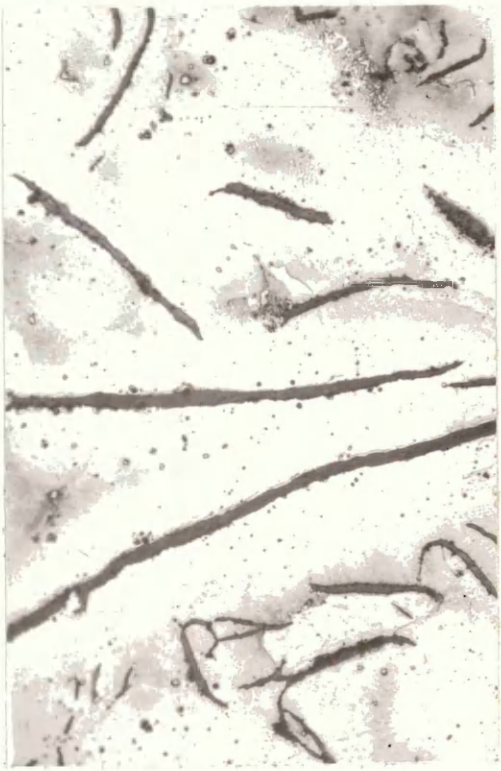
b) RG719 - 0.056% Ti

Figure 22 (Cont'd)



x10 x100

c) RG770 - 0.10% Ti



x10 x100

d) RG755 - 0.27% Ti

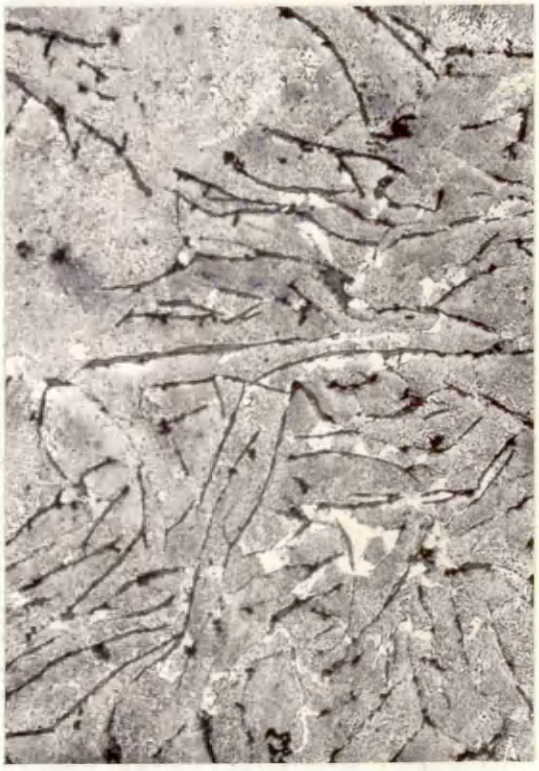
Figure 22 Micrographs Of Carbon Dioxide Treated Heavy Section Test Blocks

x10

x100

a) RG801 - 0.036% Ti, Clean Scrap

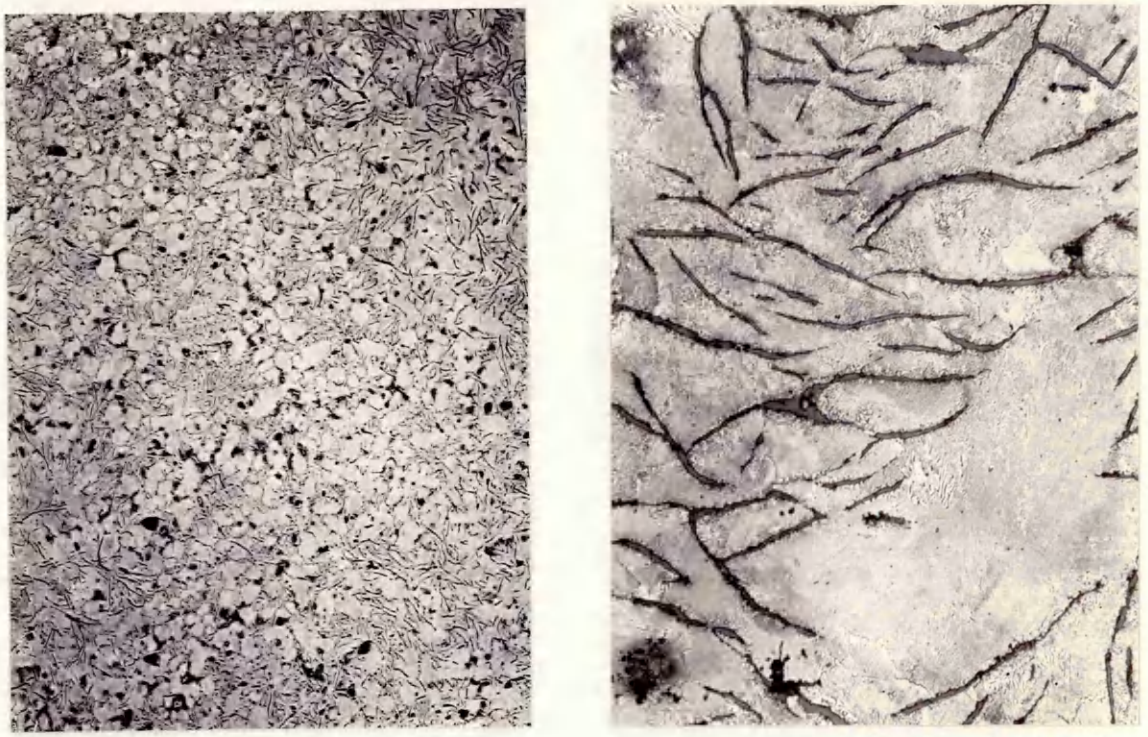
x10



x100

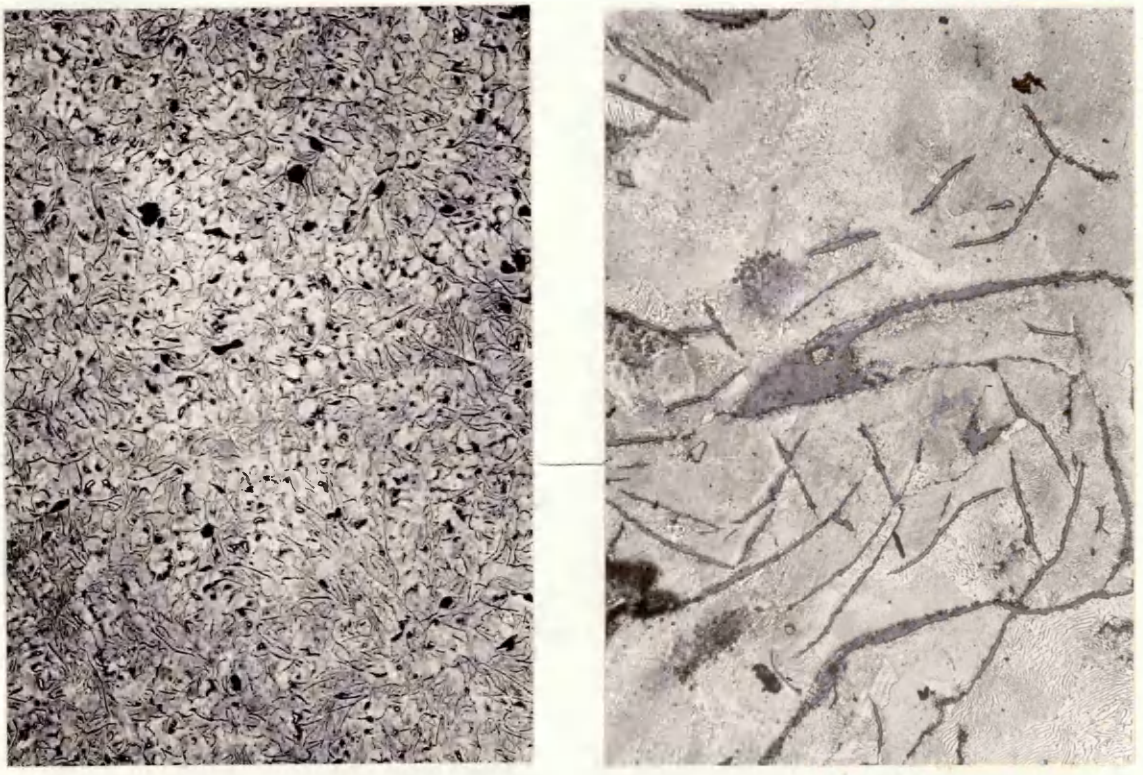
b) RG802 - 0.049% Ti, Clean Scrap

Figure 23 (Cont'd)



x10 x100

c) RG803 - 0.037% Ti, Dirty Scrap



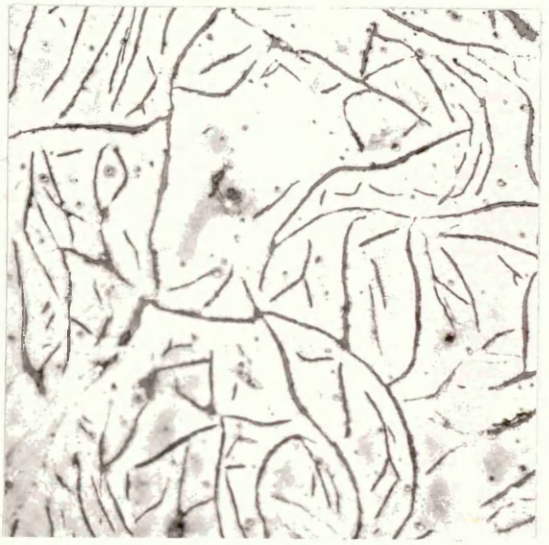
x10 x100

d) RG804 - 0.06% Ti, Dirty Scrap

Figure 23 Micrographs Of Heavy Section Test Blocks Made With Clean And Dirty Mould Scrap

Figure 24 Nitrogen Pinholing In Cast RG752A

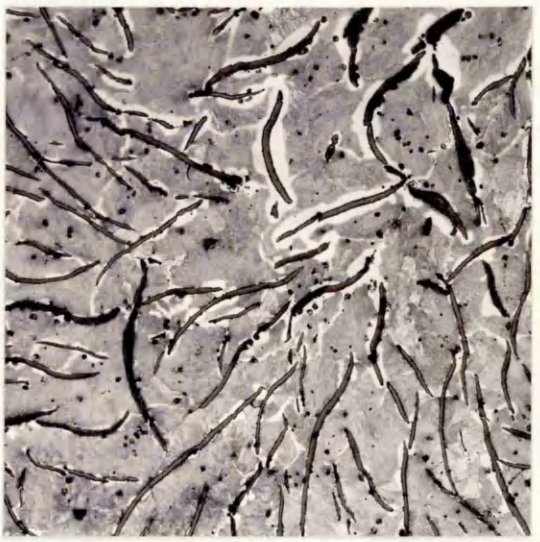
a) RG667A High NFeMn



b) RG668A N2 Gas

Figure 25 (Cont'd)

c) RG666A 0.5% $CaCN_2$

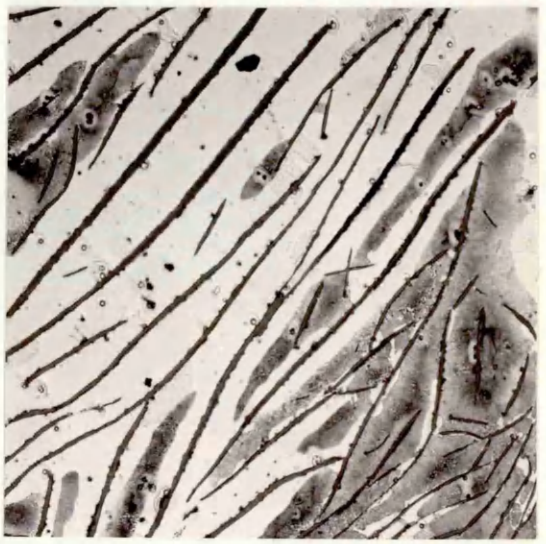


d) RG666B 0.5% Calsiloy

e) RG751A 1.0% $CaCN_2$

f) RG751B 1.0% Calsiloy

g) RG752A 2.0% $CaCN_2$



h) RG752B 2.0% Calsiloy

Figure 25 Microstructures Of Test Blocks With Various Nitrogen And Calcium Treatments

x100

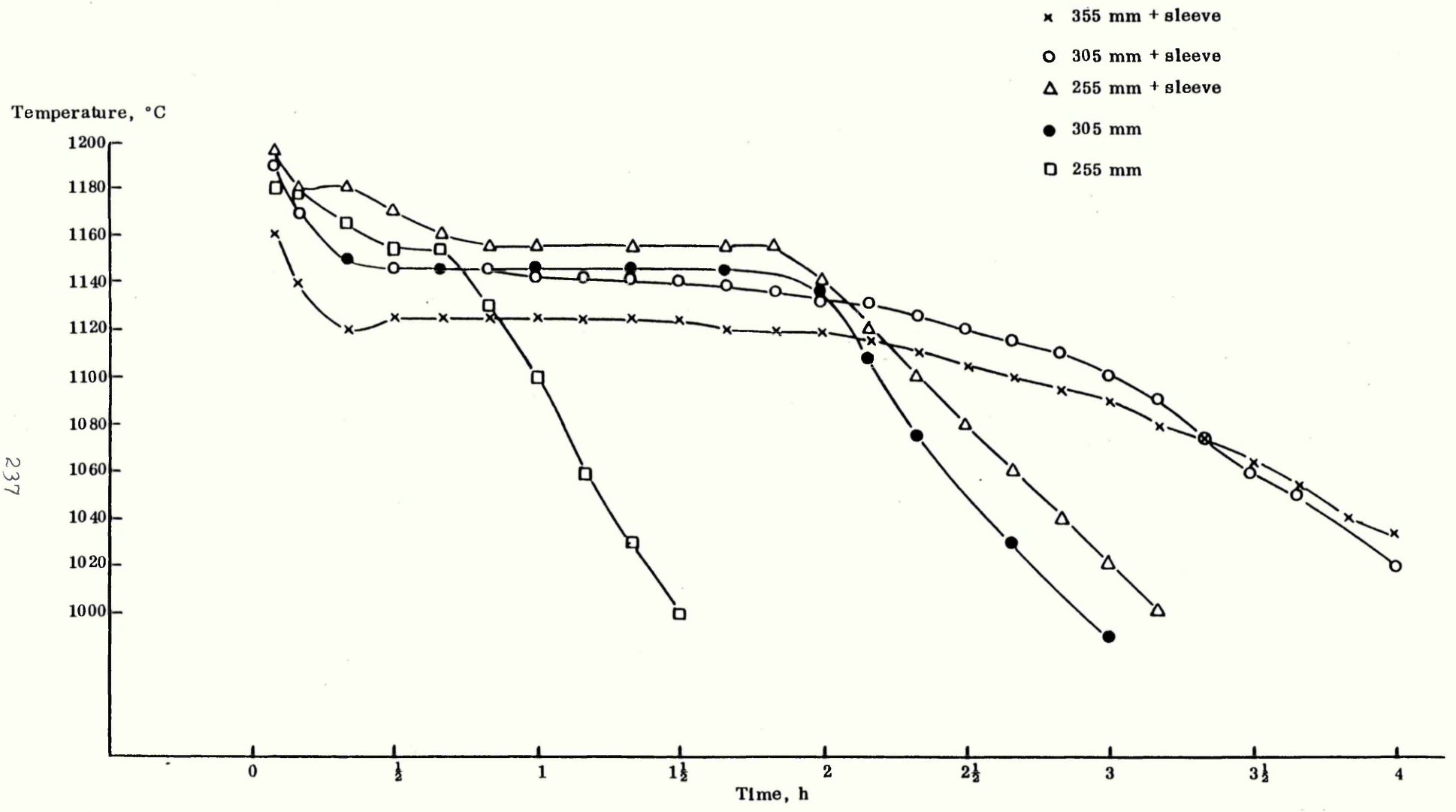


Figure 26a Solidification Times Of The Test Blocks

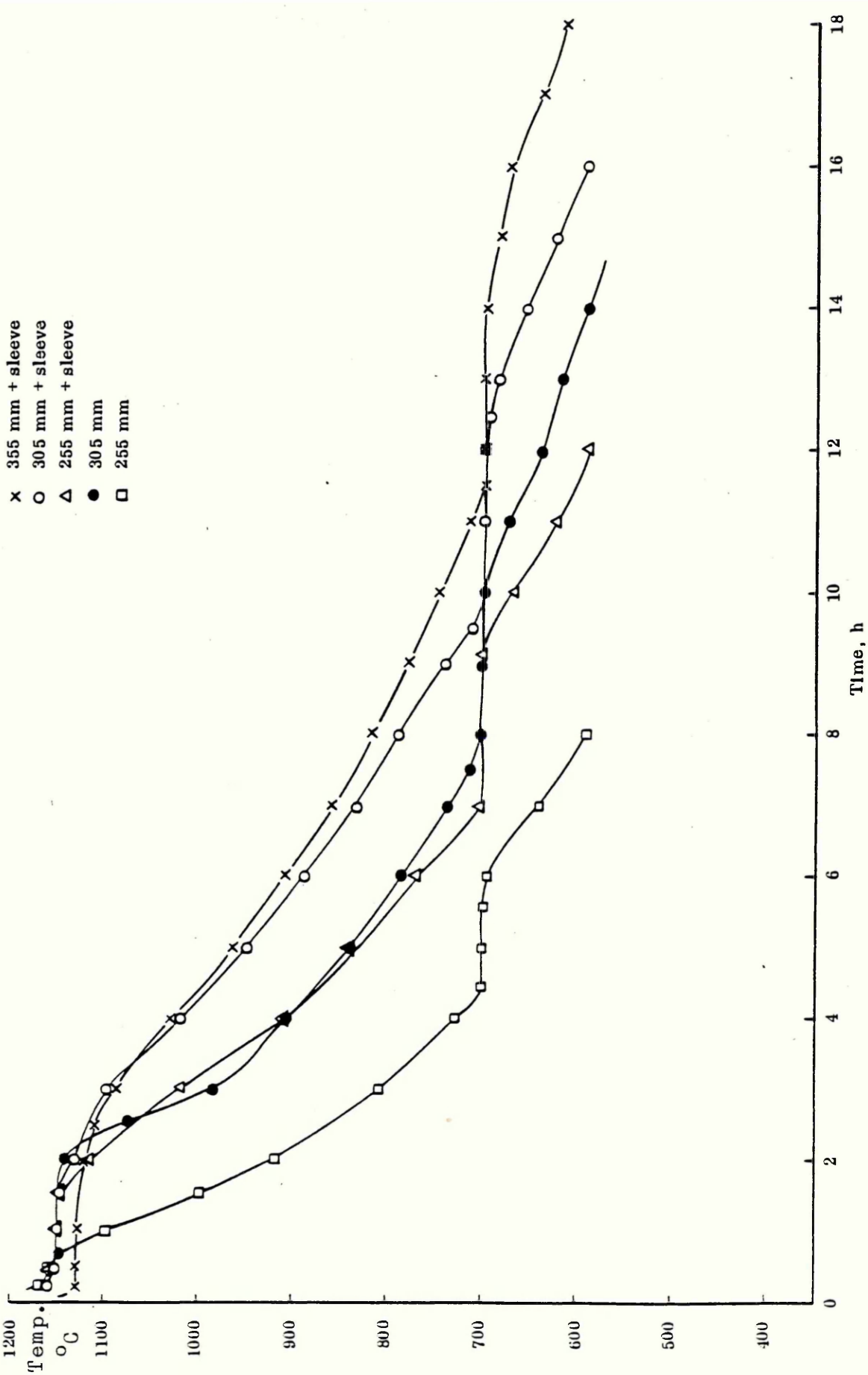


Figure 26b Cooling Times Of The Test Blocks

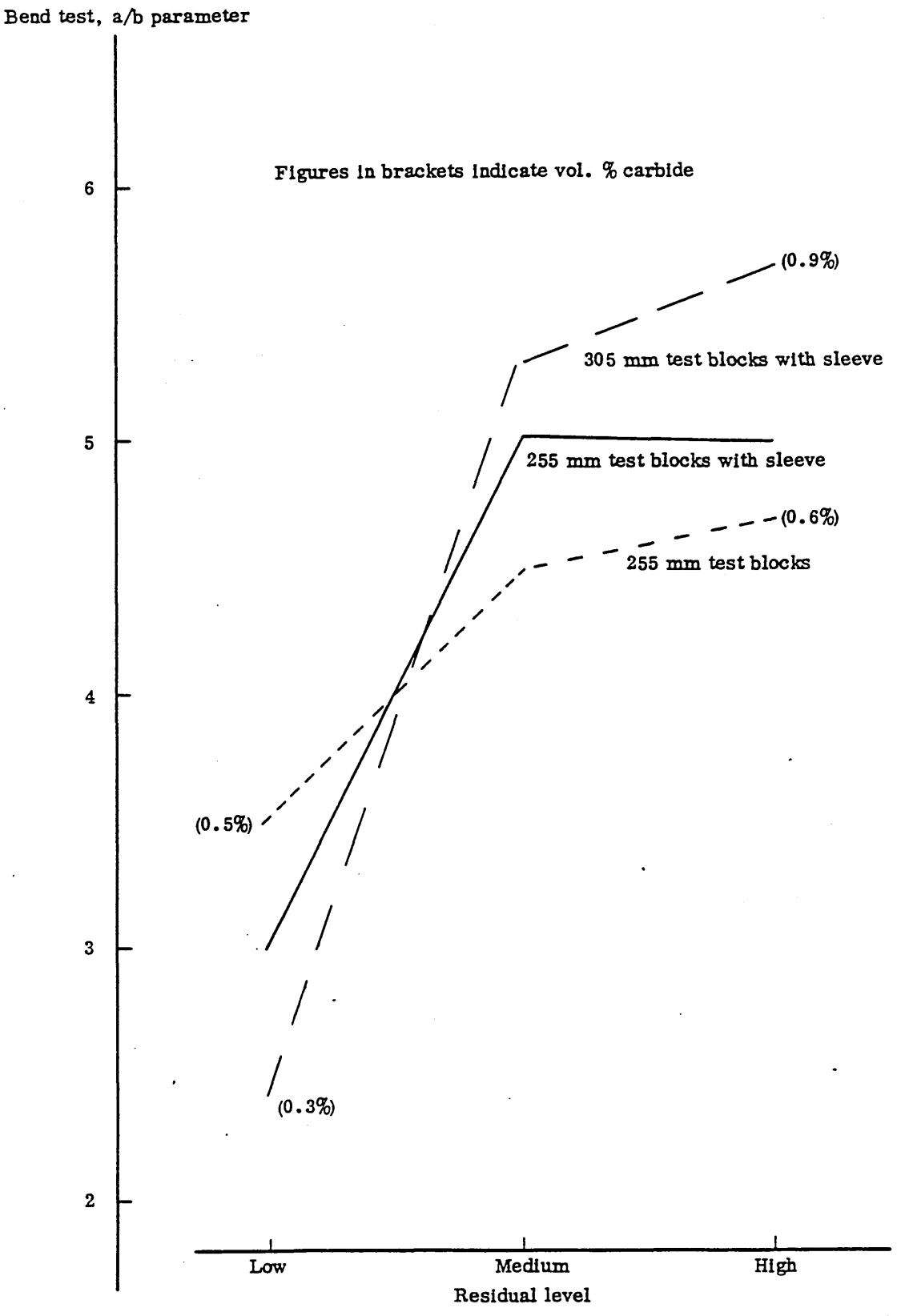
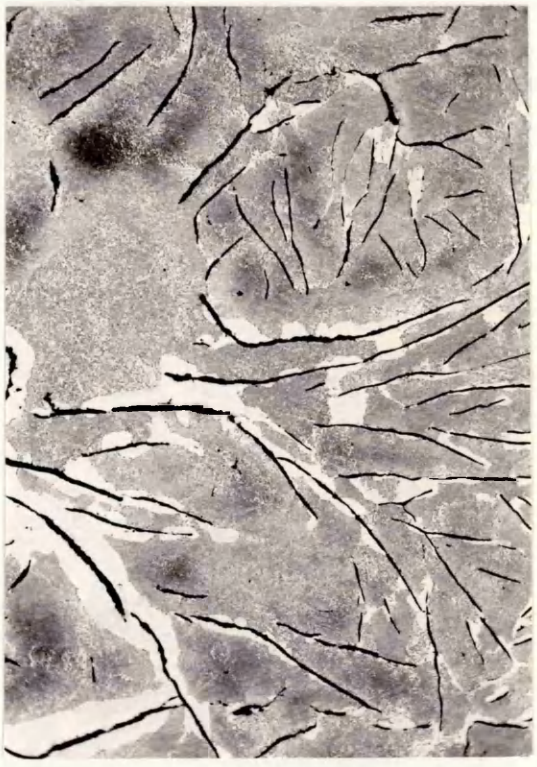


Figure 27 The Effect Of Residual Elements On Embrittlement

RG720A 255mm Block

RG720B 255mm Block With Sleeve

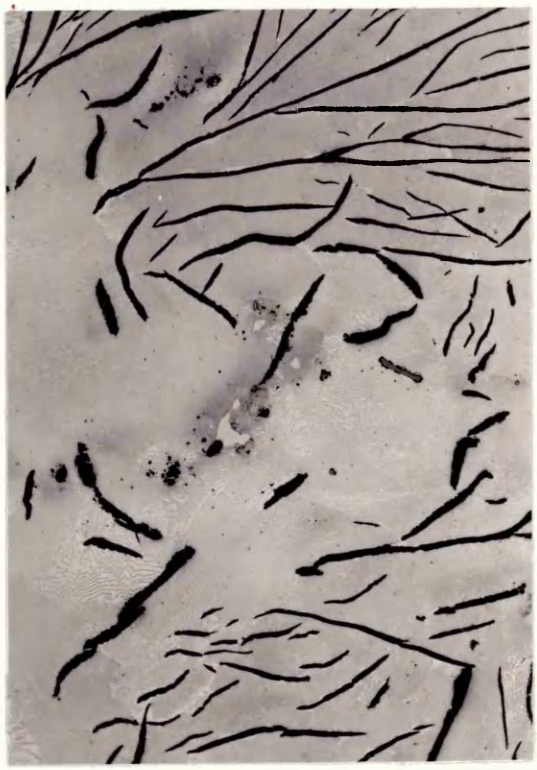
RG853 305mm Block With Sleeve



RG852 355mm Block With Sleeve

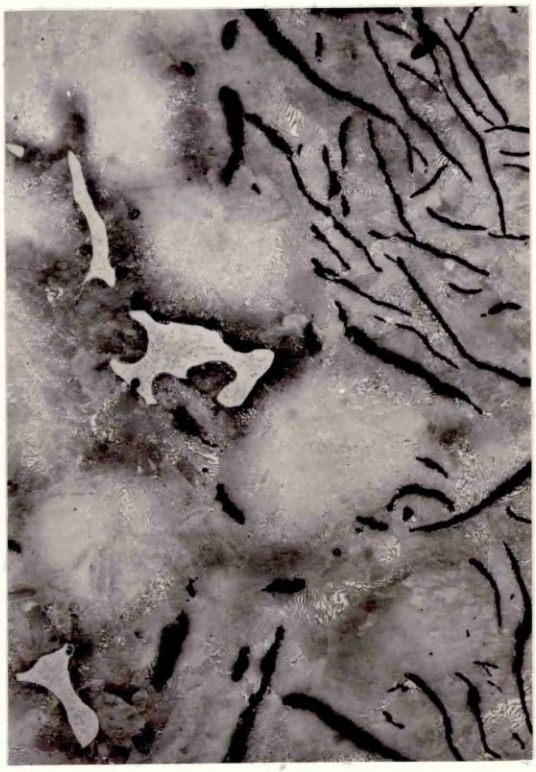
Figure 28 Microstructures Of The Low Residual Melts

x100



RG854A 255mm Block

RG854B 255mm Block With Sleeve



RG865 305mm Block With Sleeve

Figure 29 Microstructures Of The Medium Residual Melts

x100

RG867A 255mm Block

RG867A Cell Boundary Detail

RG867B 255mm Block With Sleeve

RG868 305mm Block With Sleeve

Figure 30 Microstructures Of The High Residual Melts

x100

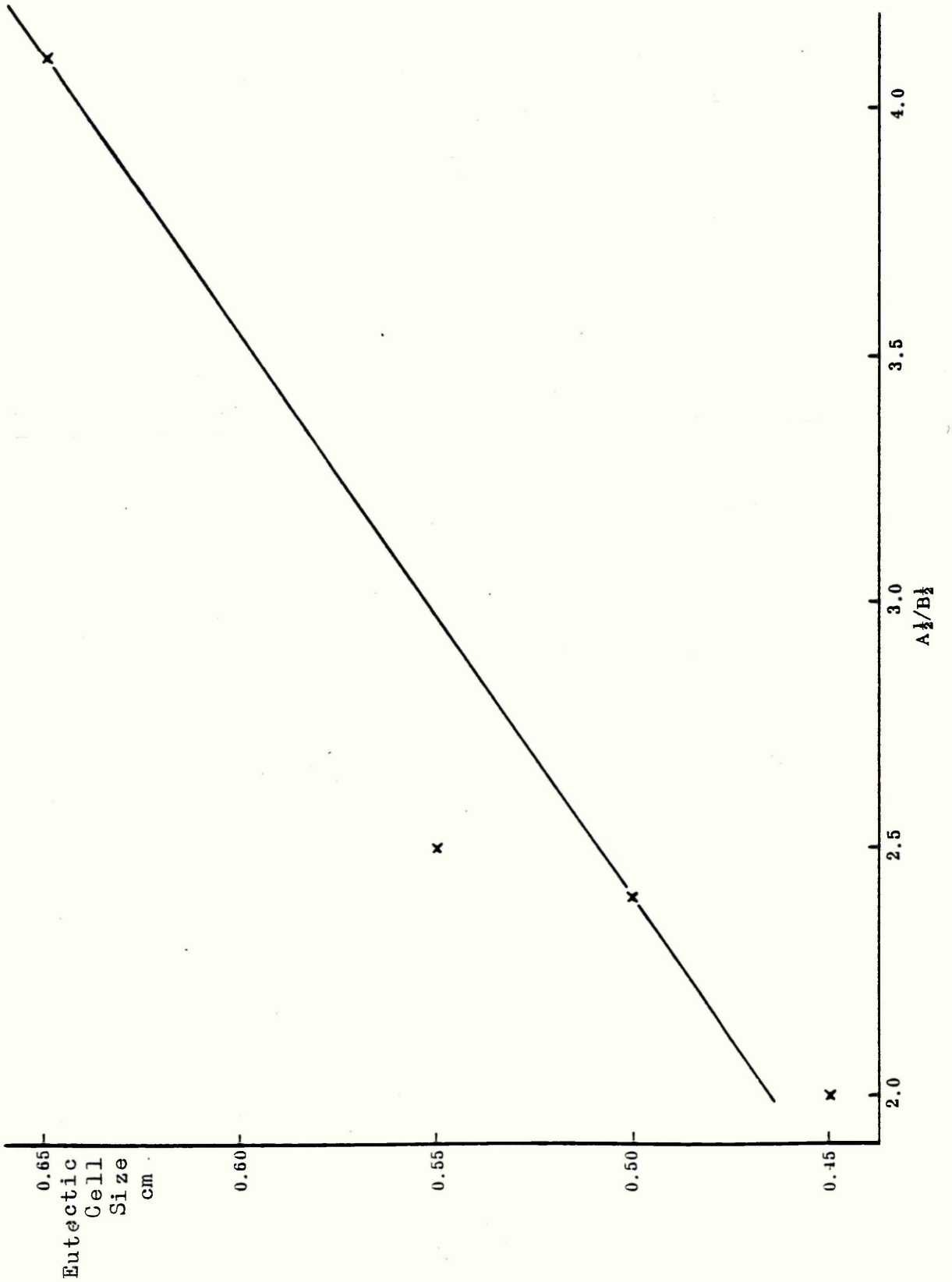


Figure 31 The Effect Of Eutectic Cell Size On Bend Test Properties

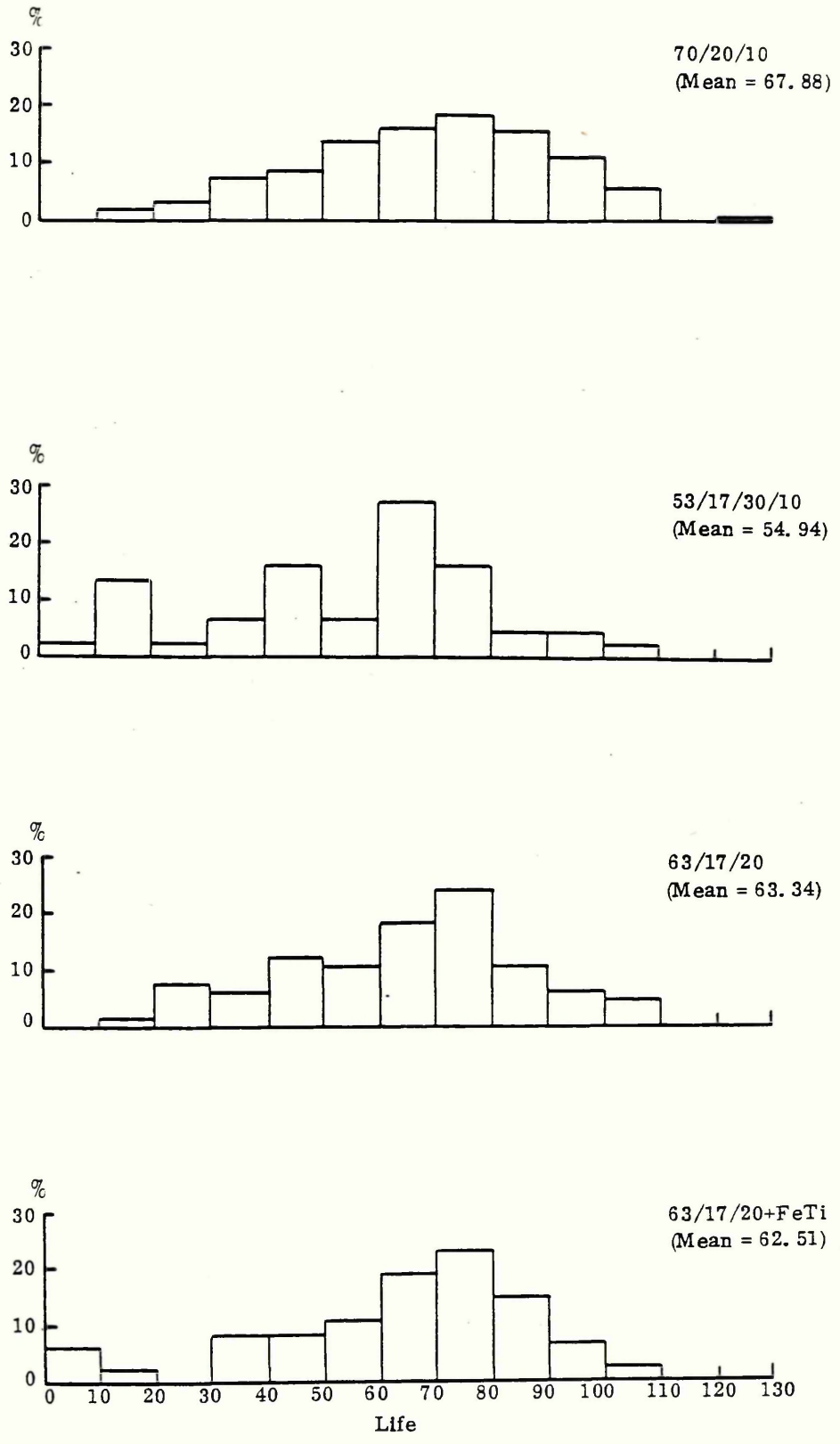


Figure 32 Distribution Of Lives For Fullwood 48-Type Moulds With Cupola Burden

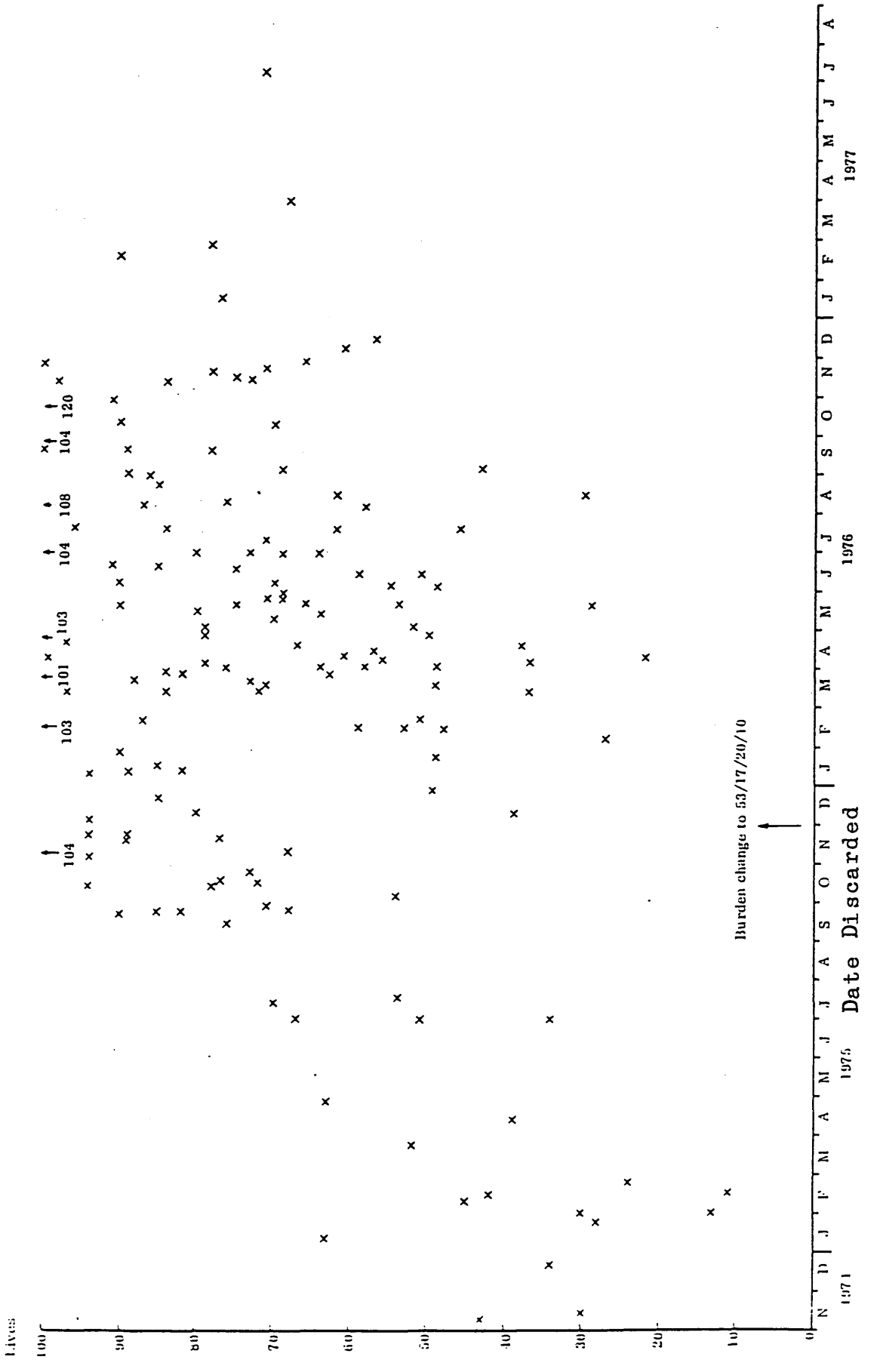


Figure 33 Life V Discard Date For 48-Type Moulds (70/20/10)

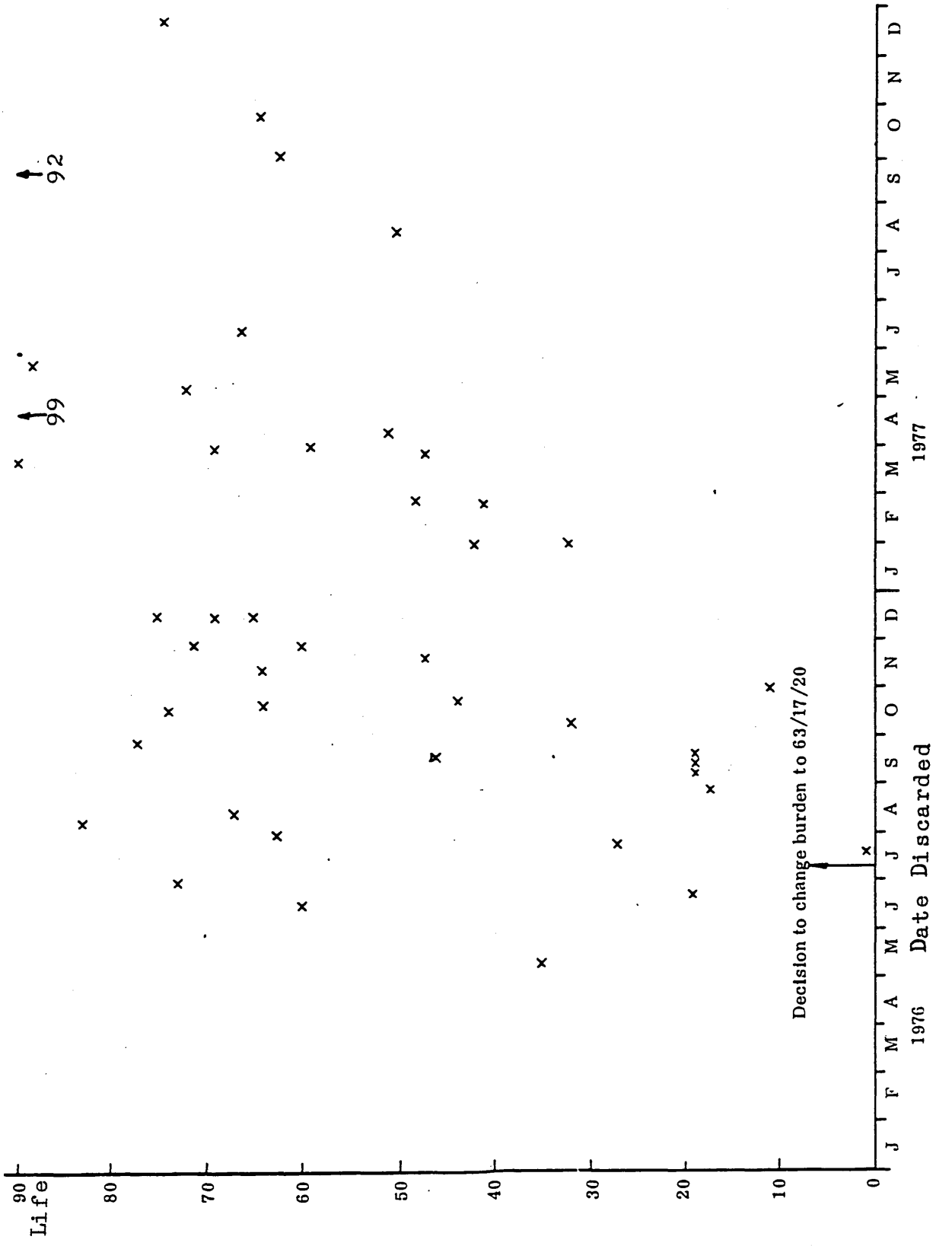


Figure 34 Life V Discard Date For 48-Type Moulds (53/17/20/10)

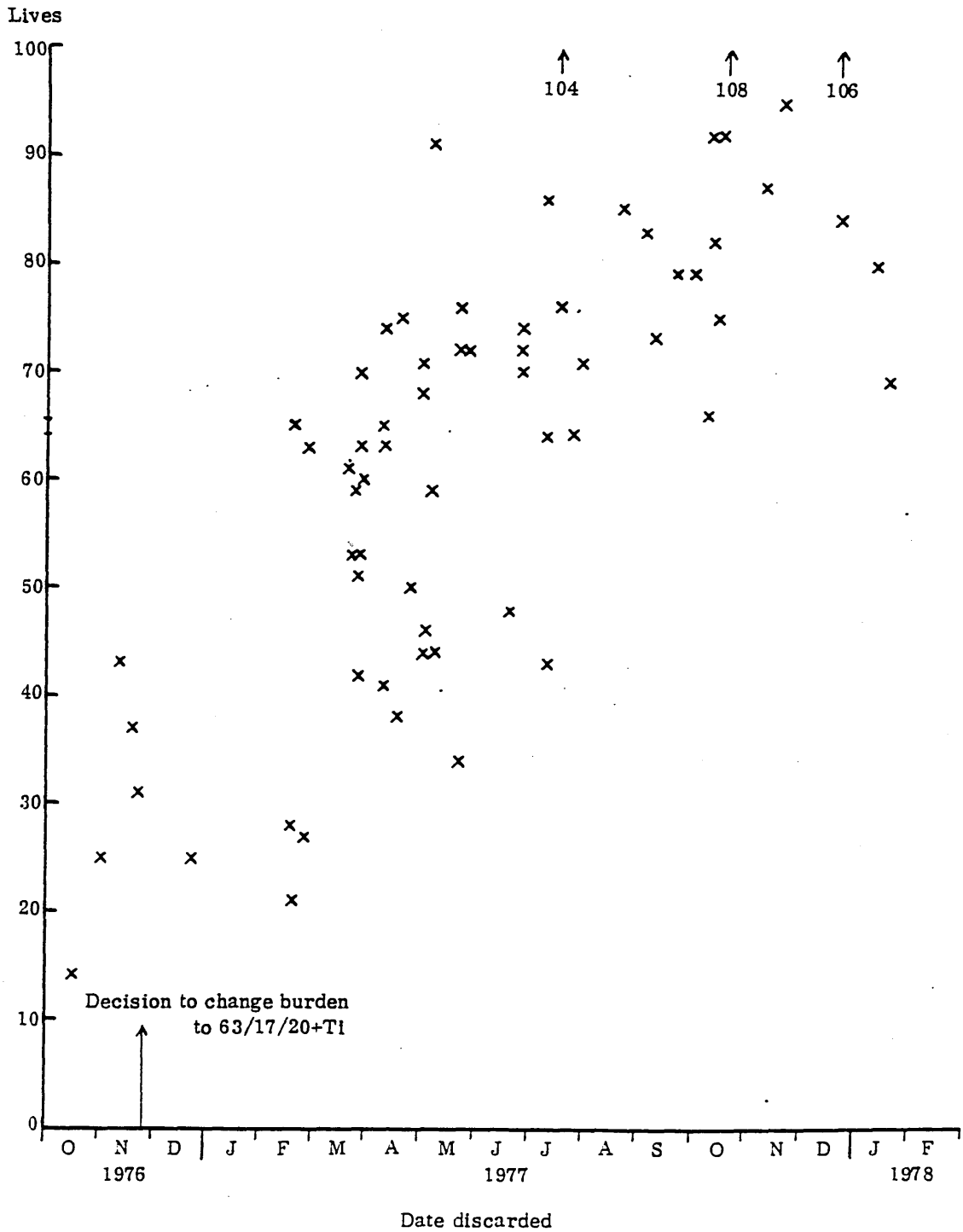


Figure 35 Life V Discard Date For 48-Type Moulds (63/17/20)

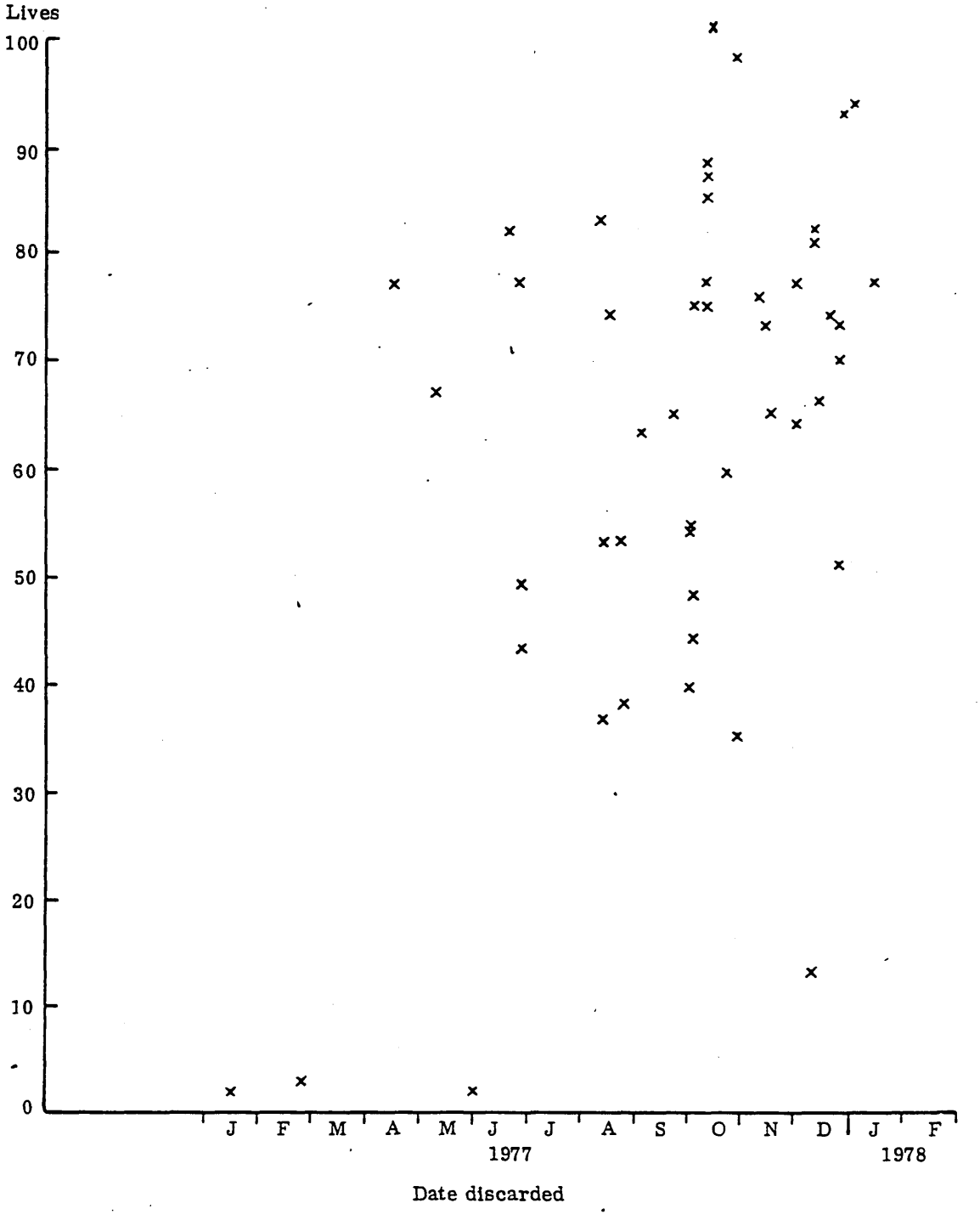
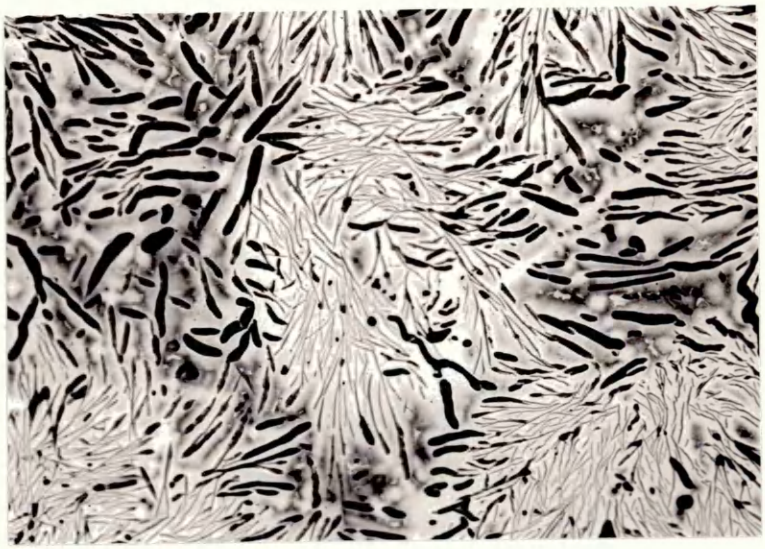


Figure 36 Life V Discard Date For 48-Type Moulds (63/17/20 + Ti)



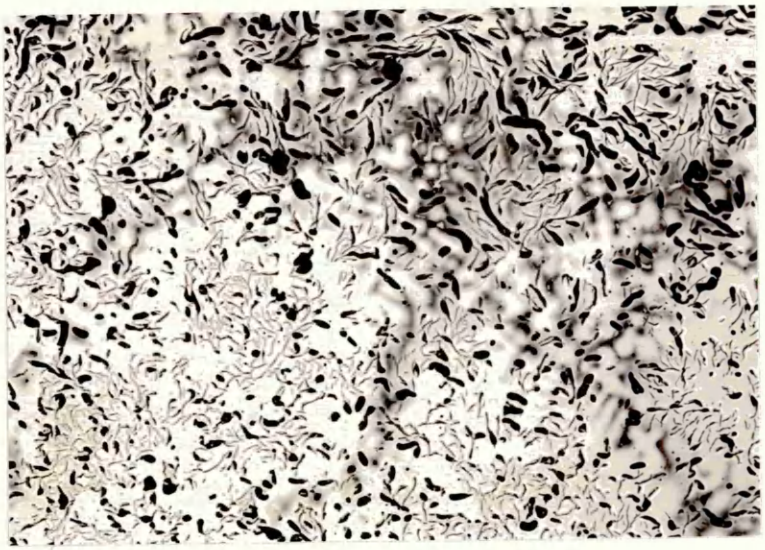
x100

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
4.05	1.35	0.035	-	0.74	-	3.64	1240°C

Cast 3.6.74 70/20/10 Burden

Figure 37 Mould No. 48/585 - 120 Lives



x10

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.80	1.66	0.038	0.040	0.54	-	-	1260°C

Cast 19.7.74 Distington Manufacture

Figure 38 Mould No. 48/758 - 65 Lives

x10

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.90	1.62	0.047	0.051	0.72	0.021	3.70	1220°C

Cast 10.12.75 53/17/20/10 Burden

Figure 39 Mould No. 48/774 - 15 Lives

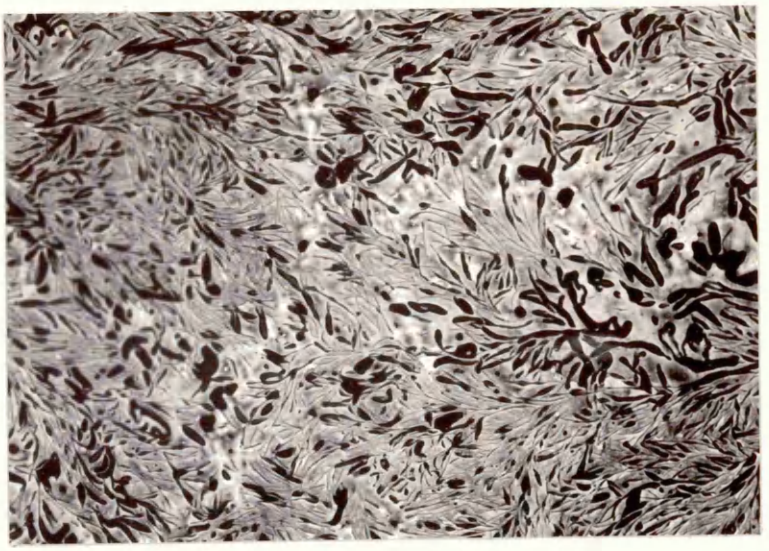
x10

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.95	1.60	0.039	0.048	0.80	0.010	3.58	1230°C

Cast 5.2.76 53/17/20/10 Burden

Figure 40 Mould No. 48/792 - 38 Lives



x10

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
4.00	1.47	0.035	-	0.85	0.032	3.60	1230°C

Cast 12.2.76 53/17/20/10 Burden

Figure 41 Mould No. 48/793 - 24 Lives

x10

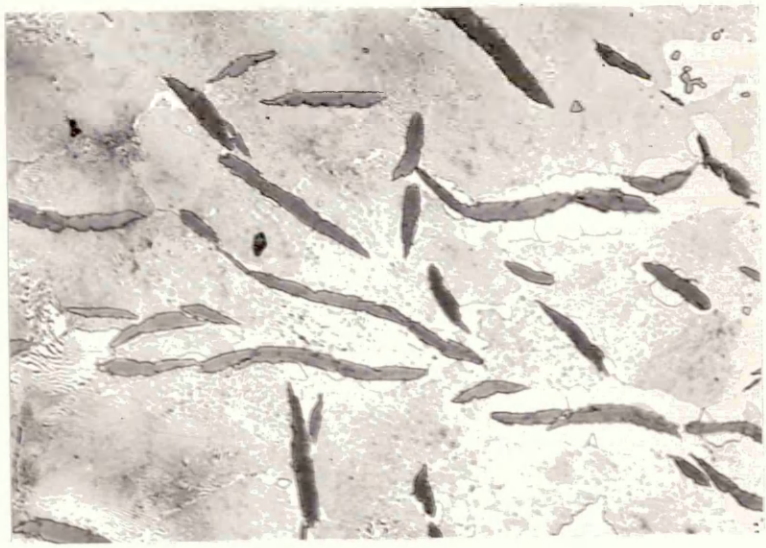
x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.82	1.48	0.071	-	-	0.019	-	1230°C

Cast 27.5.76 53/17/20/10 Burden

Figure 42 Mould No. 48/839 - 15 Lives

x10



x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.96	1.41	0.065	-	0.71	-	-	1245°C

Cast 23.5.76 53/17/20/10 Burden

Figure 43 Mould No. 48/840 - 16 Lives

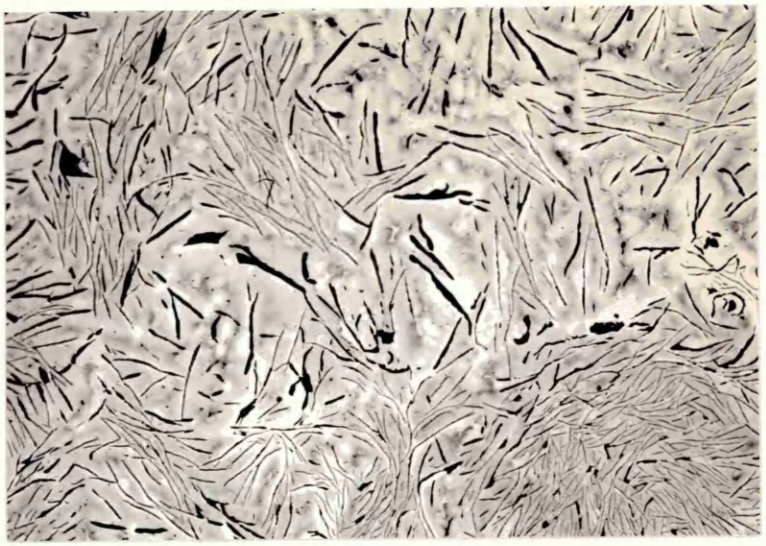
x10

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.91	1.30	0.067	-	0.79	0.043	-	1280°C

Cast 30.6.76 63/17/20 Burden

Figure 44 Mould No. 48/879 - 13 Lives



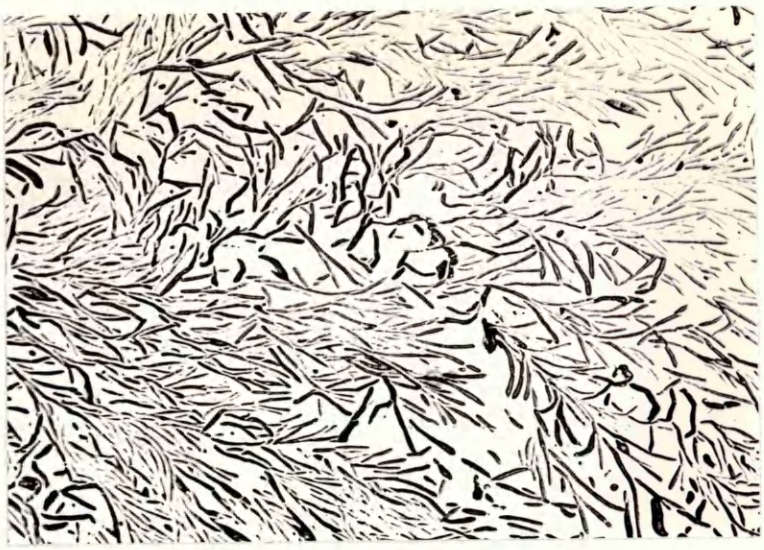
x10

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.92	1.47	0.054	-	0.73	0.053	-	1250°C

Cast 10.11.76 63/17/20+Ti Burden

Figure 45 Mould No. 48/023 - 77 Lives



x10

x100

C	Si	S	P	Mn	Ti	USV	Casting Temperature
3.82	1.39	0.056	-	0.78	0.057	3.30	1280°C

Cast 30.11.76 63/17/20+Ti Burden

Figure 46 Mould No. 48/059 - 2 Lives

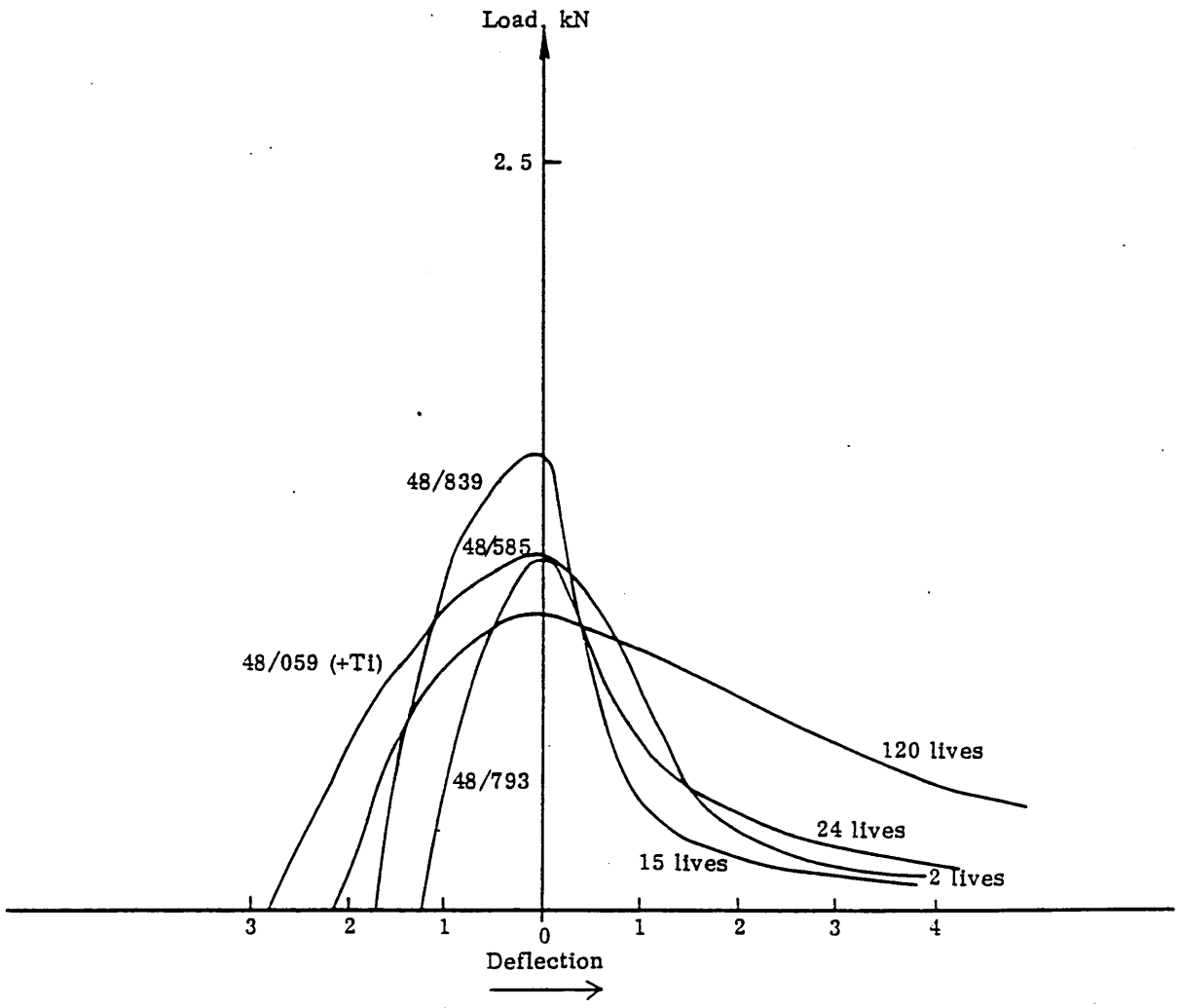


Figure 47 Bend Test Curves Of Flake Graphite Moulds

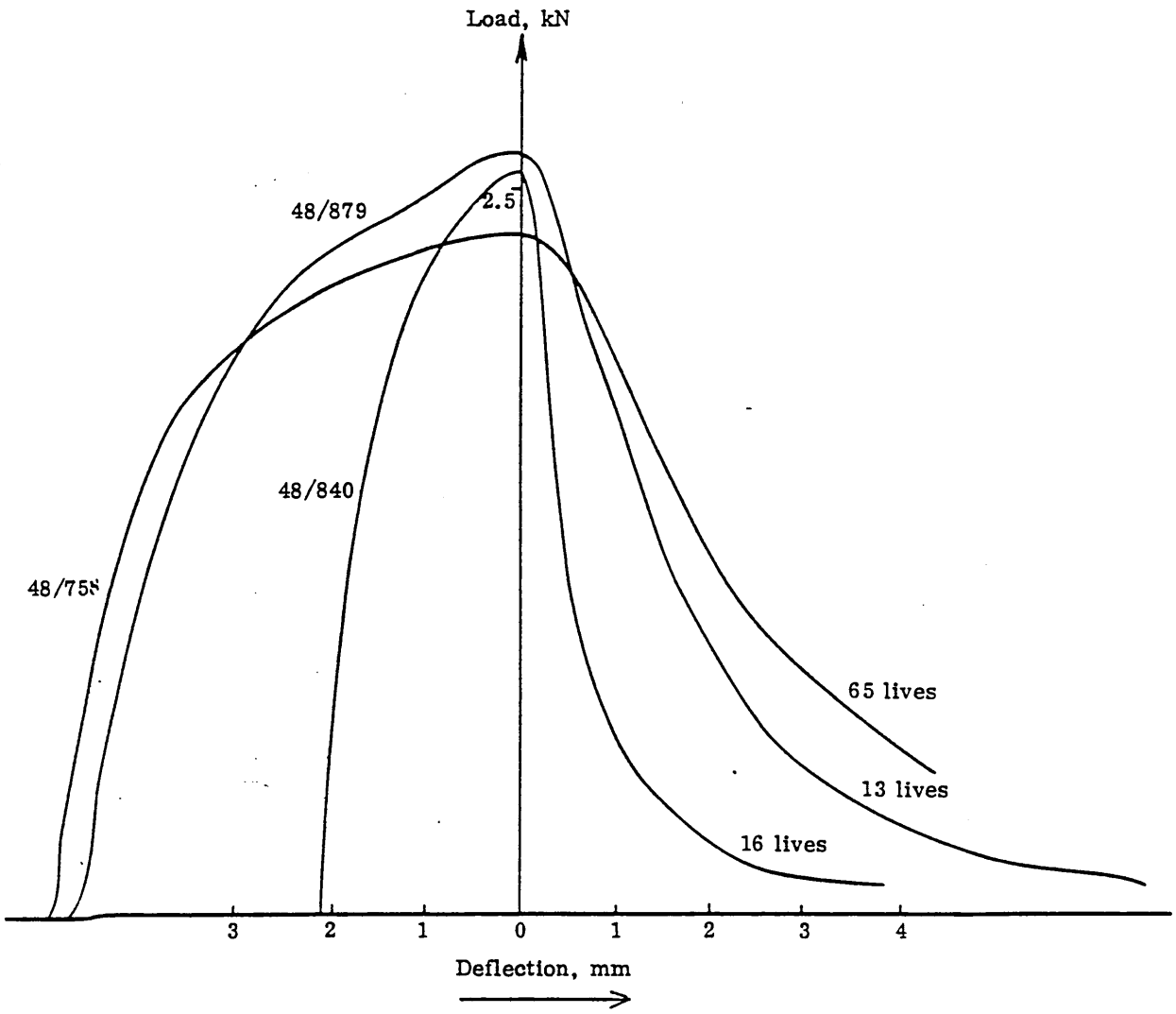
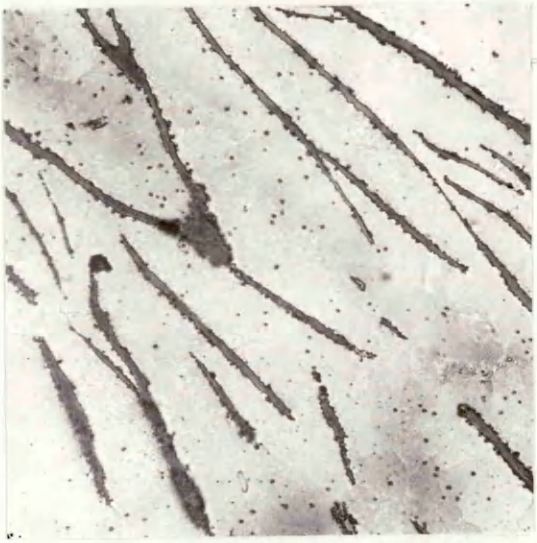


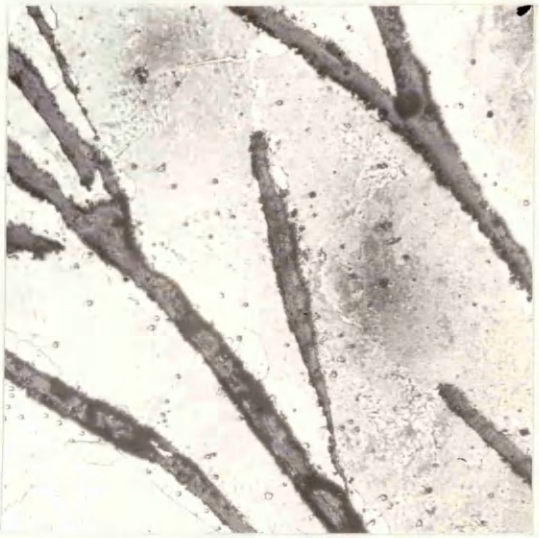
Figure 48 Bend Test Curves Of Compacted Graphite Moulds

Mould 79/75 No.1

Mould 105/101 No.2

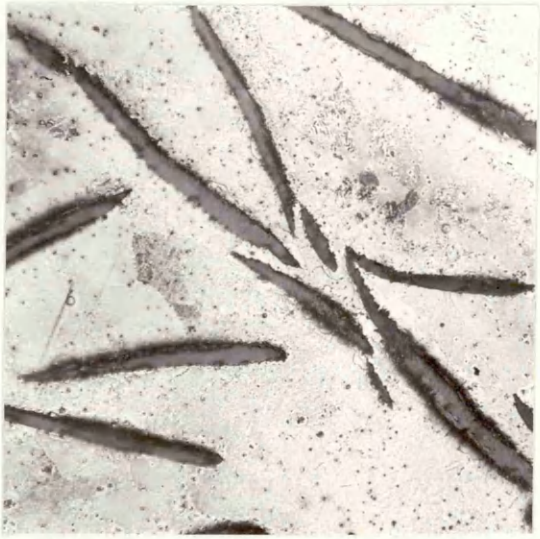


Mould 71/68x16 No.1



Mould 91/86 No.2

Mould 105/101 No.4



Mould 71/68x16 No.2

Figure 49 Typical Microstructures Of River Don Forging Moulds

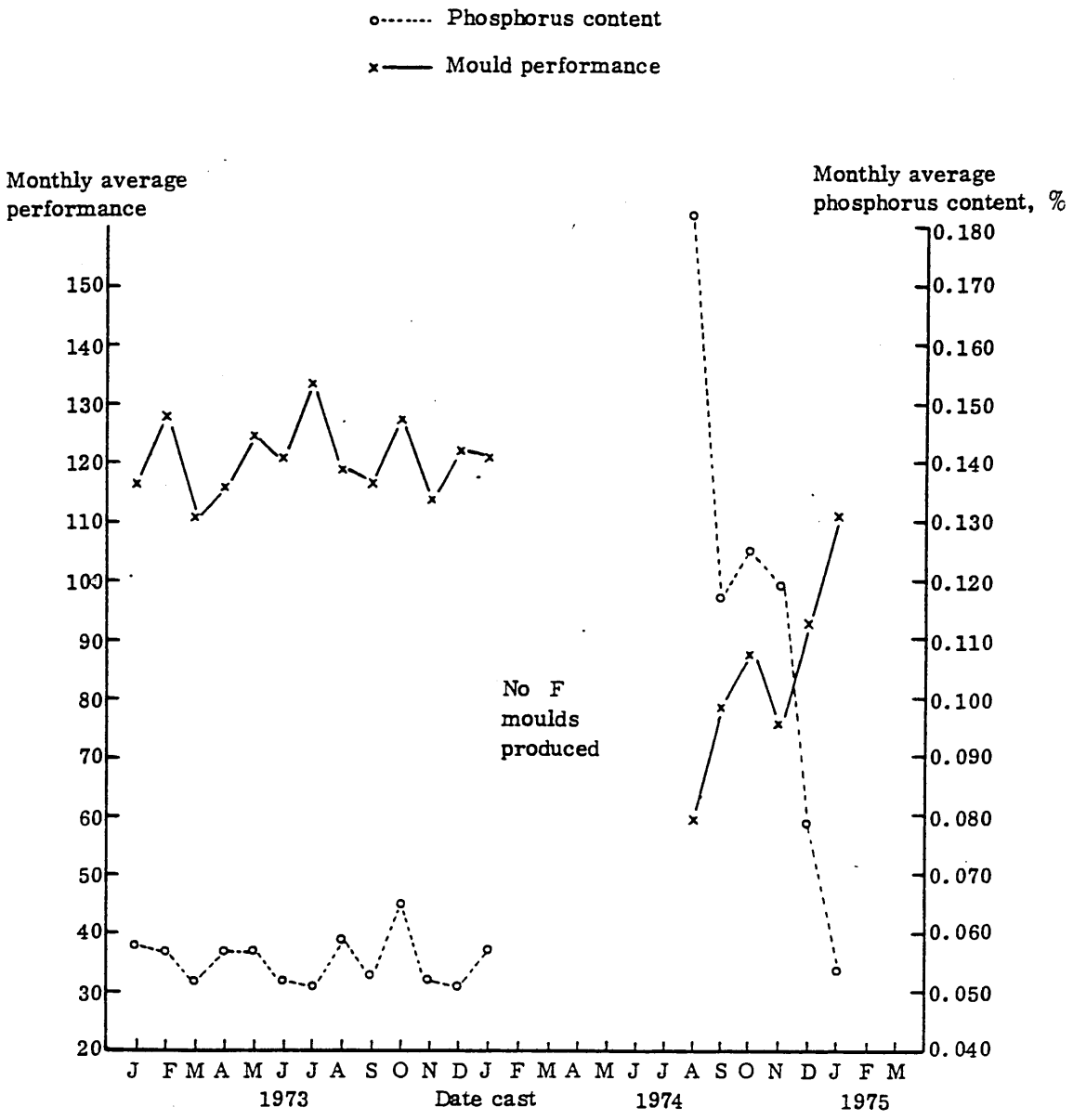


Figure 50 The Monthly Average Performance And Phosphorus Content Of Renishaw F-Type Moulds

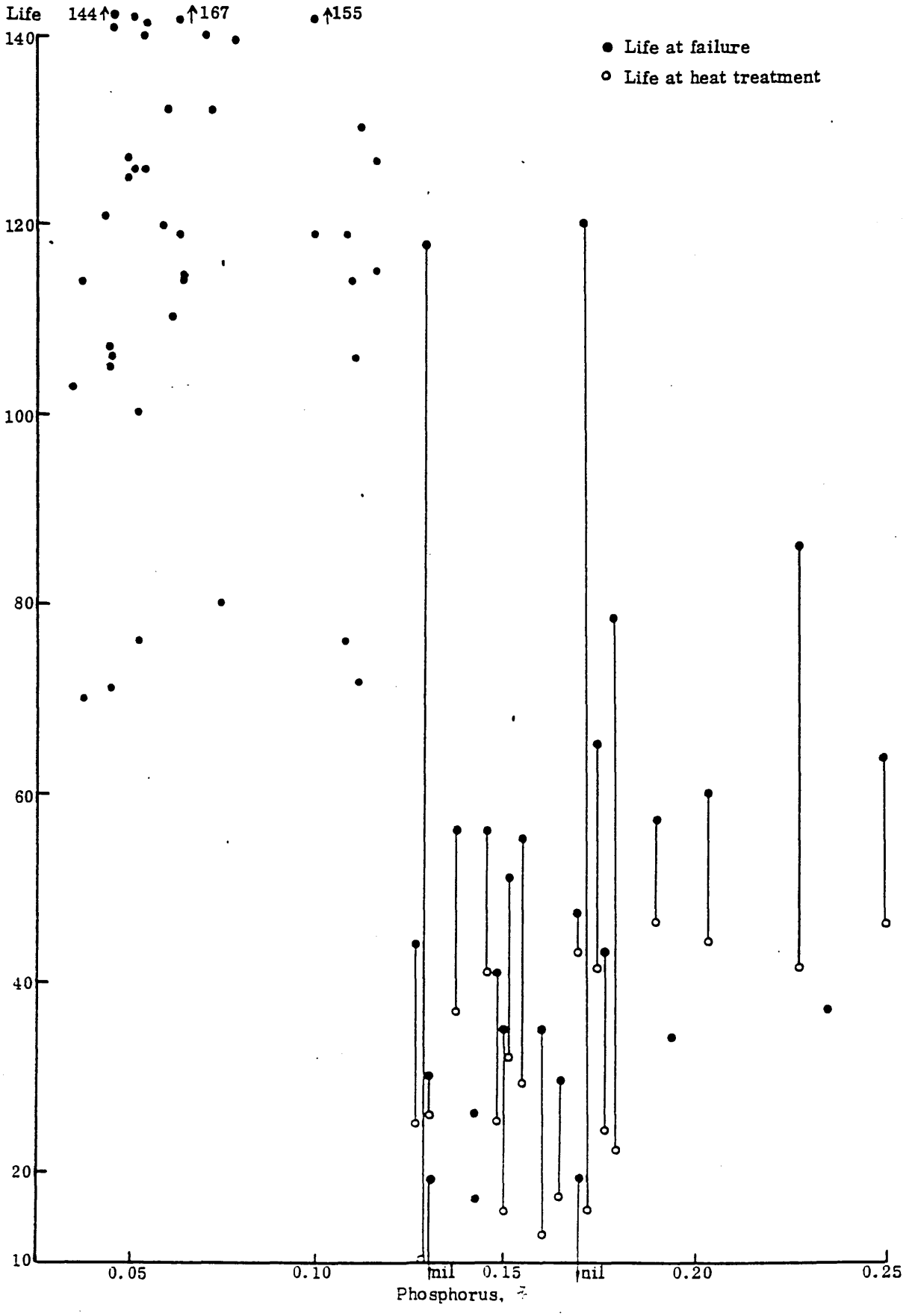


Figure 51 The Effect Of Phosphorus Content On The Performance Of F-Type Moulds Manufactured Aug To Dec 1974 At Renishaw

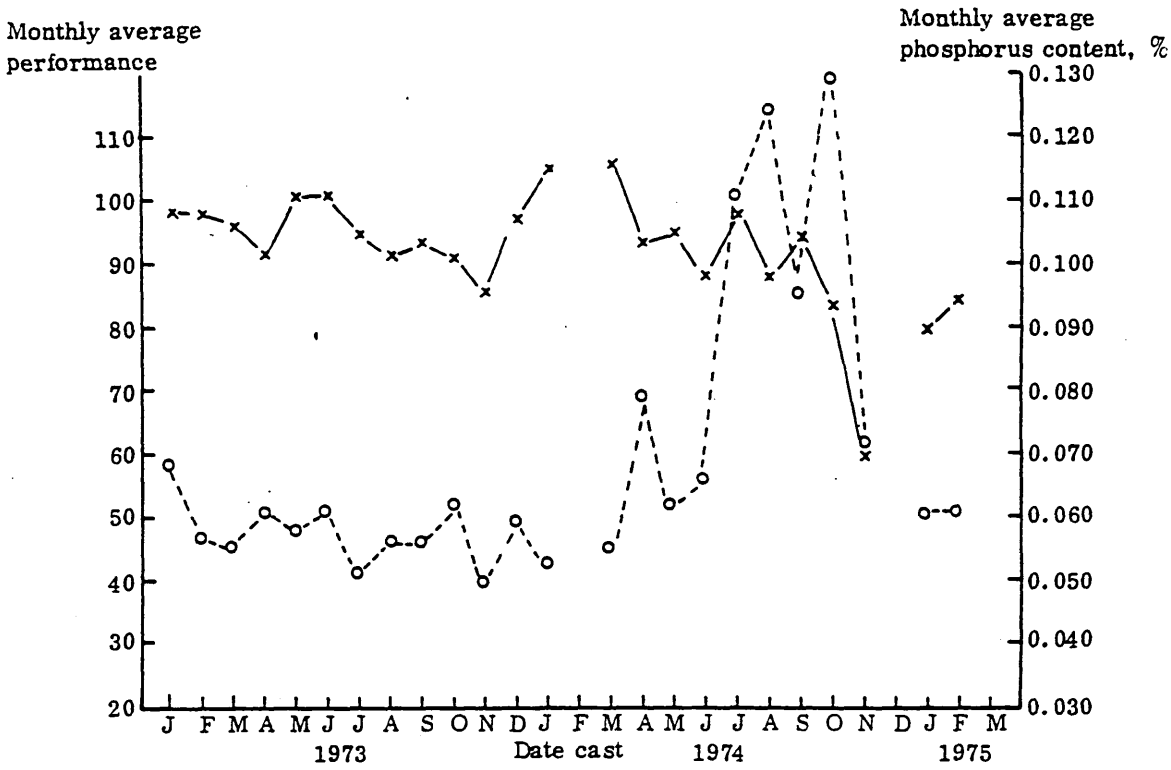


Figure 52 The Monthly Average Performance And Phosphorus Content Of Renishaw 610WB Moulds

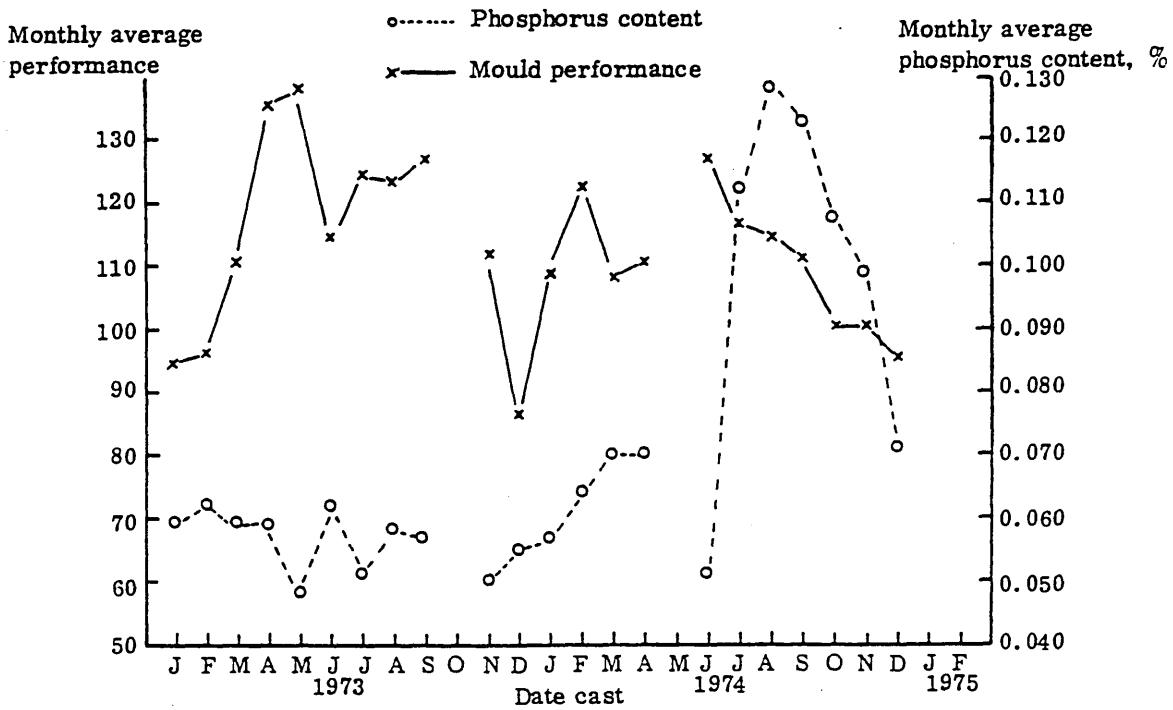


Figure 53 The Monthly Average Performance And Phosphorus Content Of Renishaw B120 Moulds

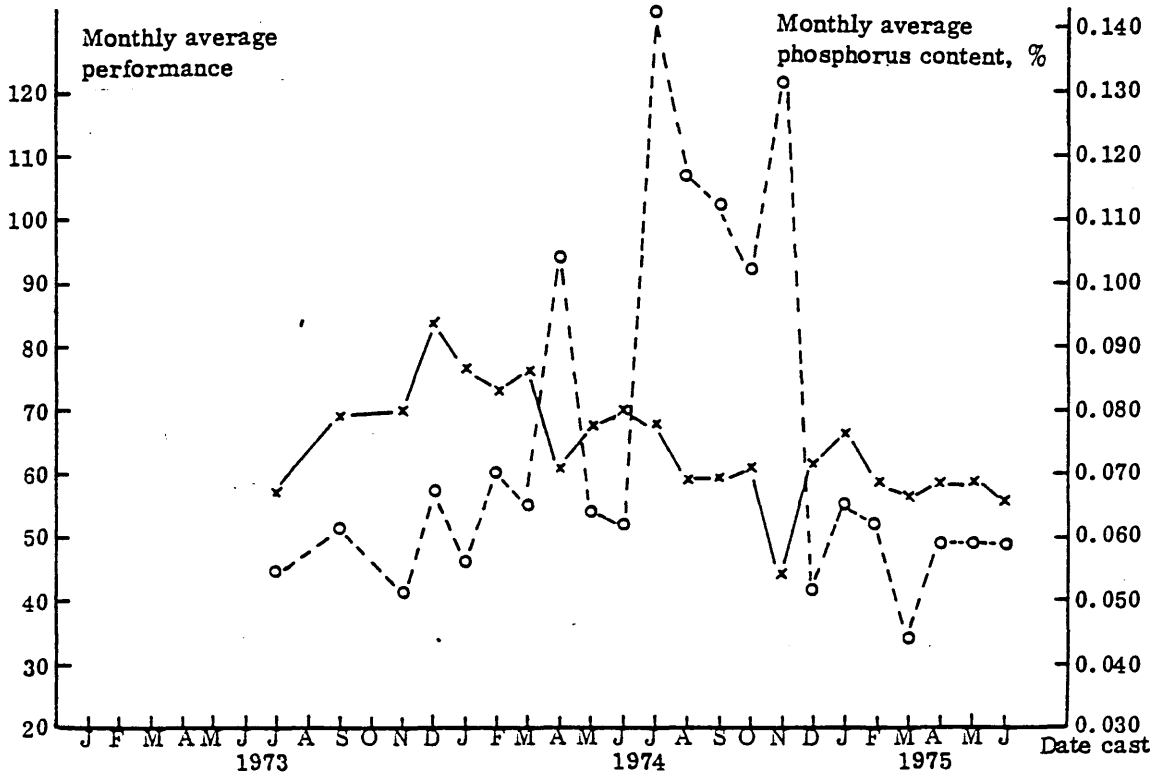


Figure 54 The Monthly Average Performance And Phosphorus Content Of Renishaw WEU100 Moulds

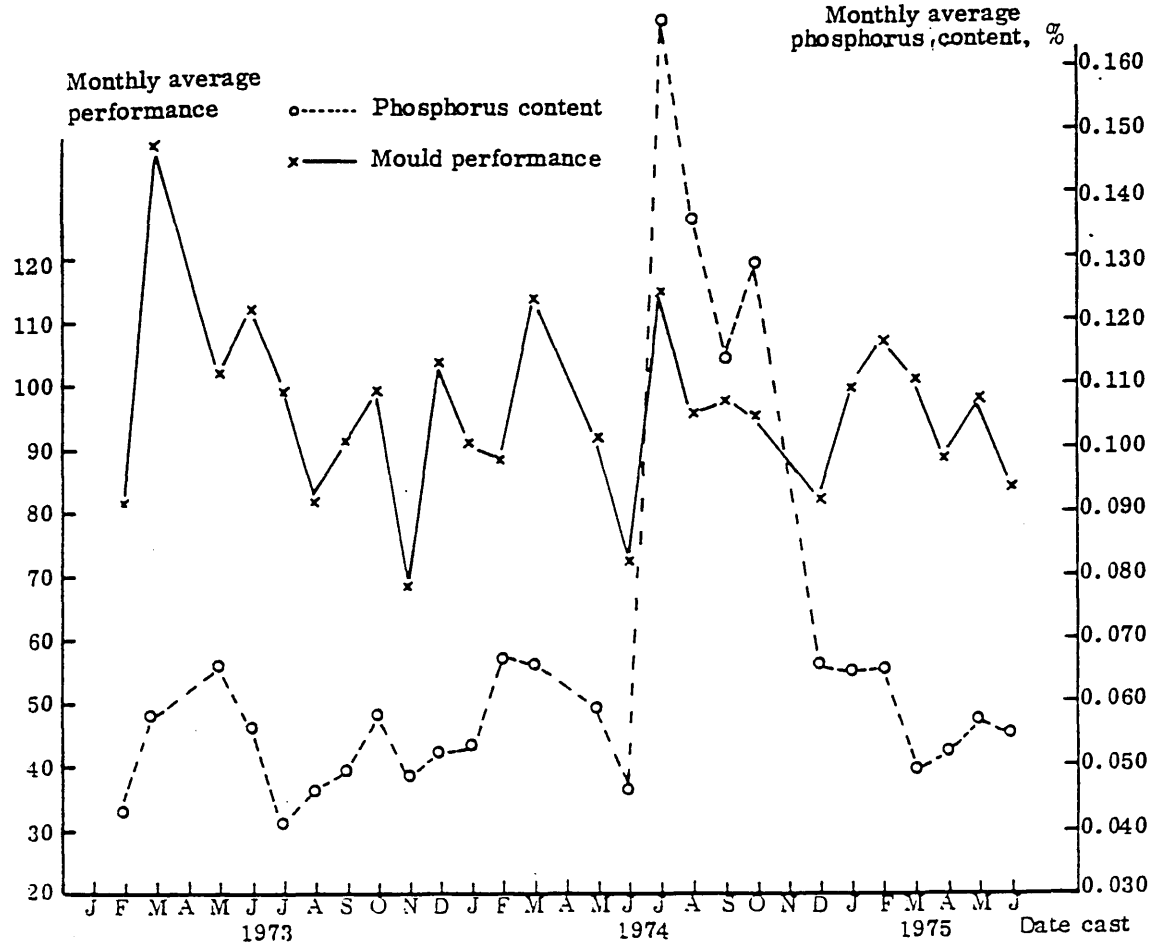


Figure 55 The Monthly Average Performance And Phosphorus Content Of Renishaw NEU100 Moulds

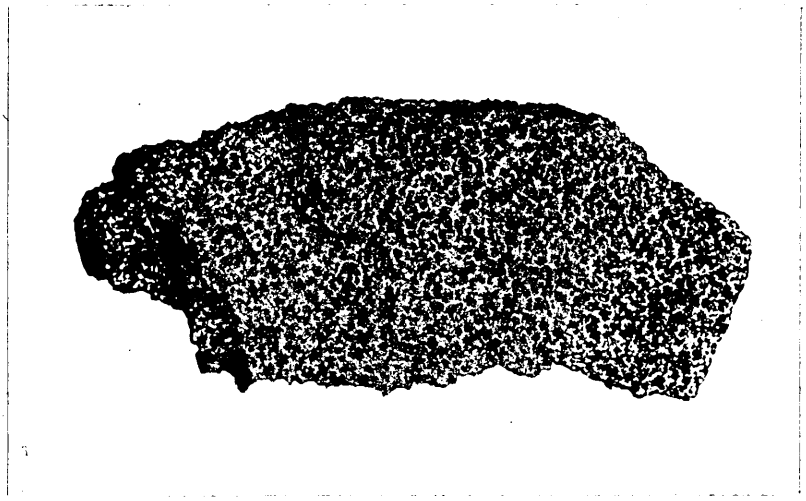


Figure 56 Fracture Surface Of Prematurely Failed High Phosphorus Mould

Figure 57 Microstructures Of Prematurely Failed High Phosphorus Moulds

x100

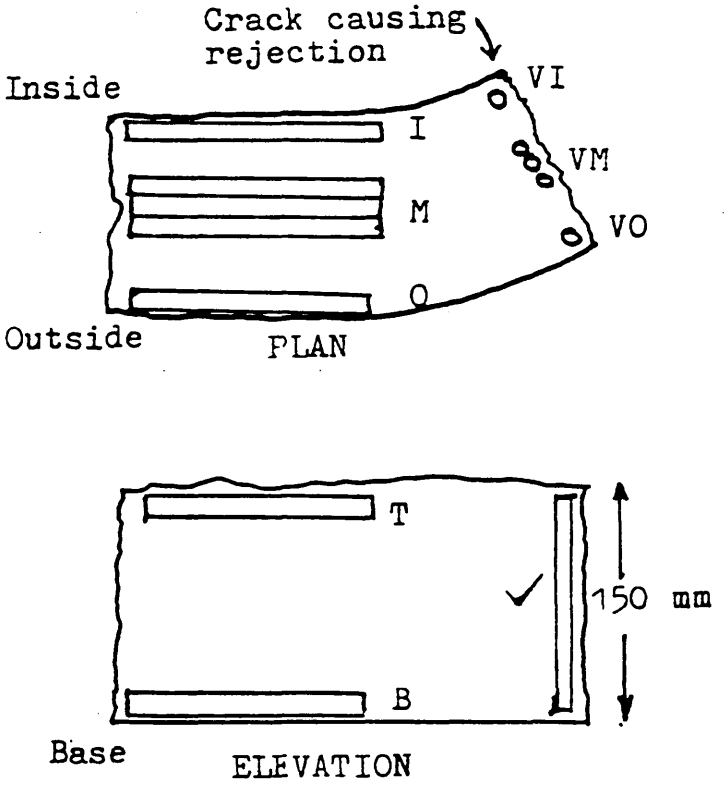


Figure 58 Sketch Showing Position Of Samples And Identification Code For WEU100 Moulds

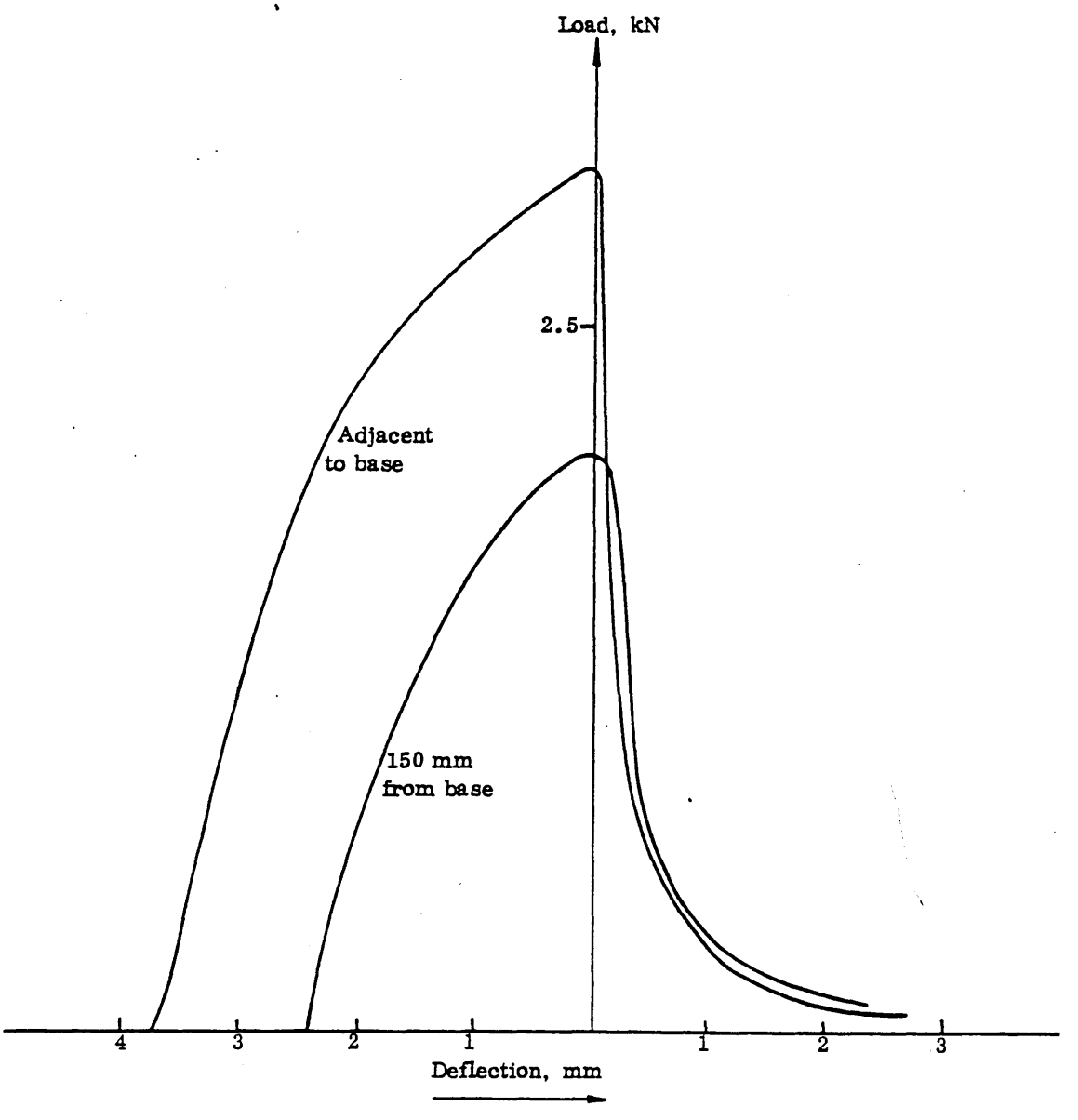
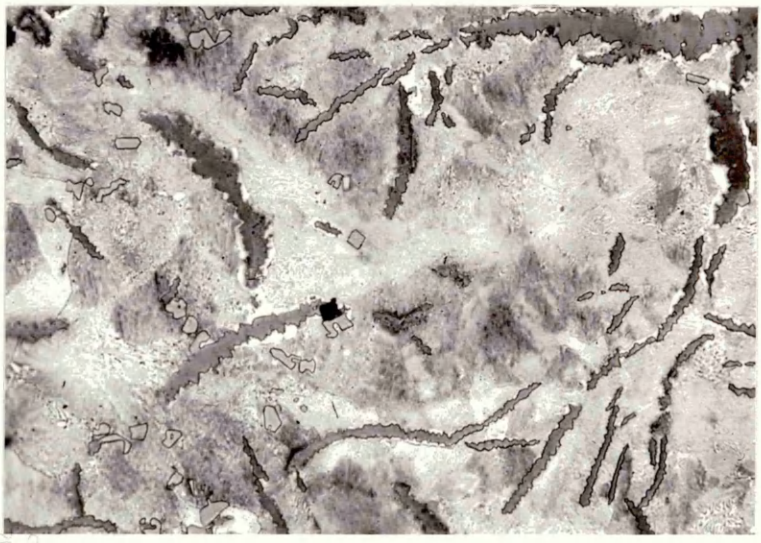


Figure 59 Typical Load-Deflection Curves For WEU100 Moulds

a) TM



b) BM

Figure 60 Microstructure Of WEU100 Mould 404

x100

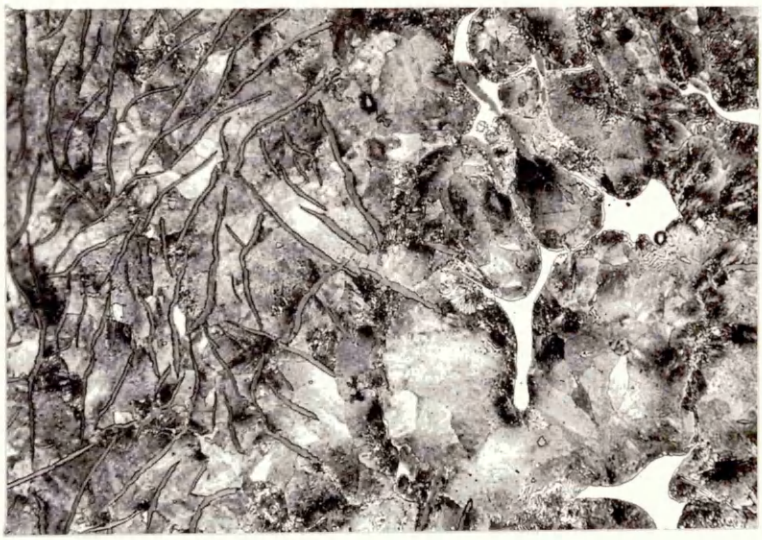


Figure 61 Microstucture Of WEU100 Mould 489 (VM) x100

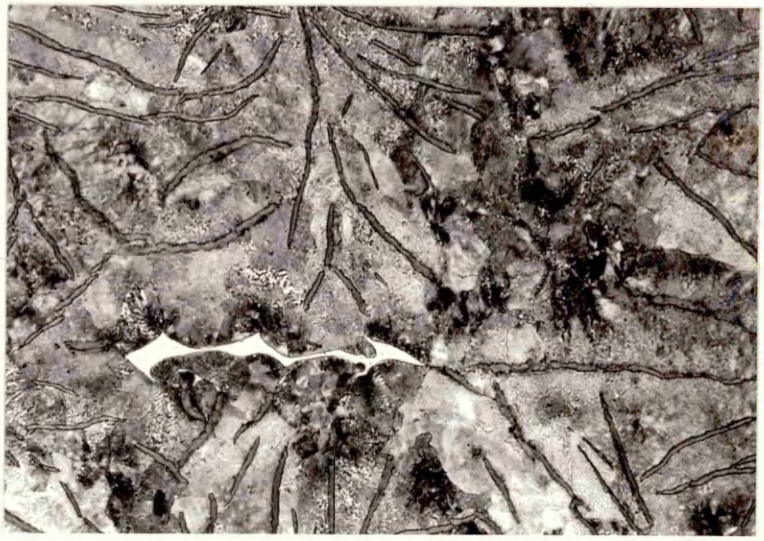


Figure 62 Microstructure Of WEU100 Mould 497 x100

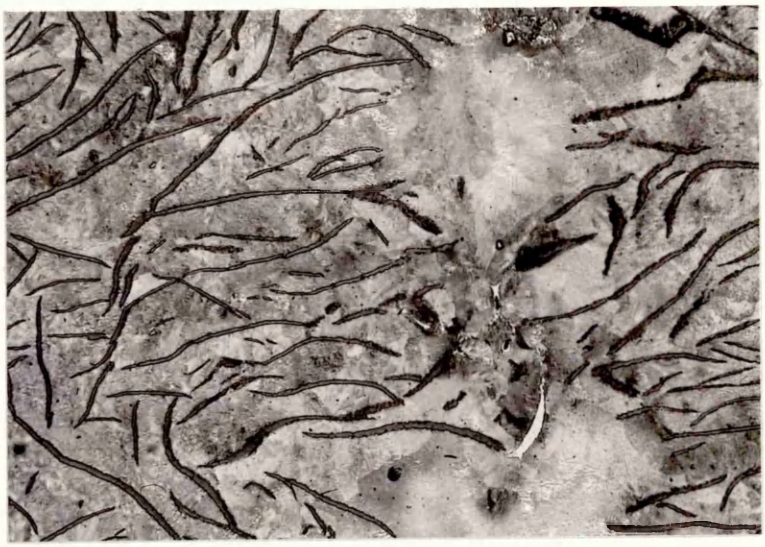


Figure 63 Microstructure Of WEU100 Mould 561 x100

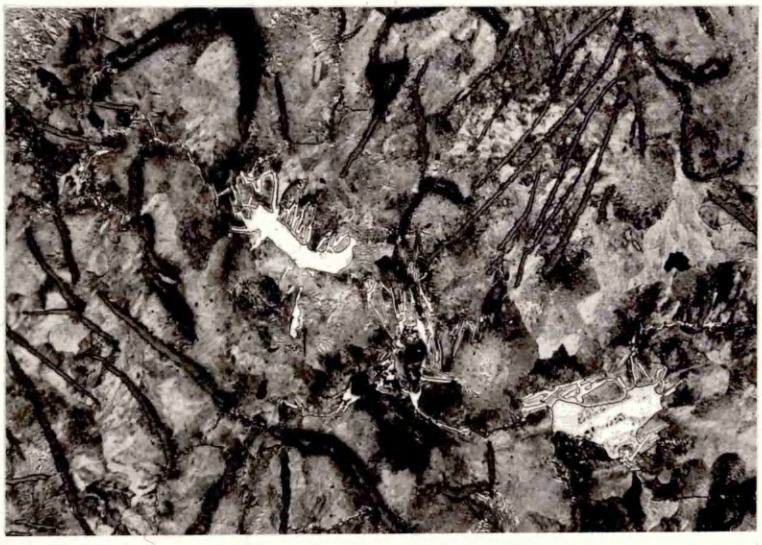


Figure 64 Microstructure Of WEU100 Mould 587 x100

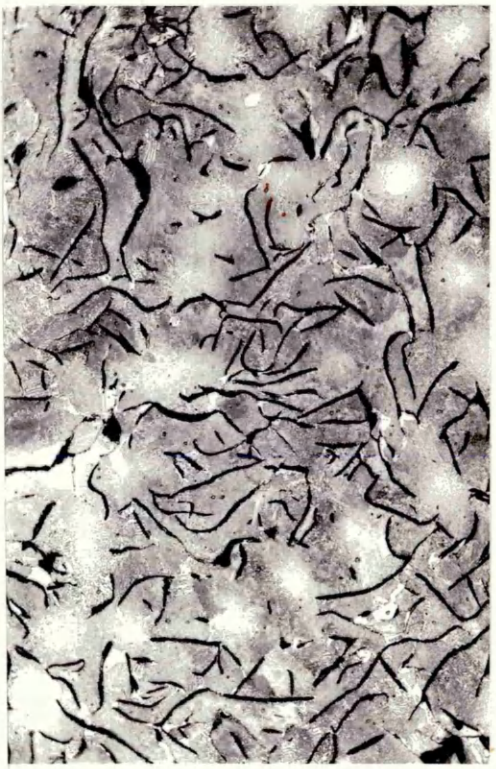
a) Renishaw Horngate 2954

b) Renishaw Lug 3348

c) Renishaw Lug 3350

Figure 65 Cont'd

d) Distington Lug 6704



e) Distington Lug 6728

Figu [illegible]cal Microstructures of Horngate and Large Samples x100

a) Dowlais

b) Landore

c) Distington

d) Fullwood

Figure 66 Cont'd

e) Craigneuk

Figure 66 Typical Microstructures of Foundry Test Block Material x100

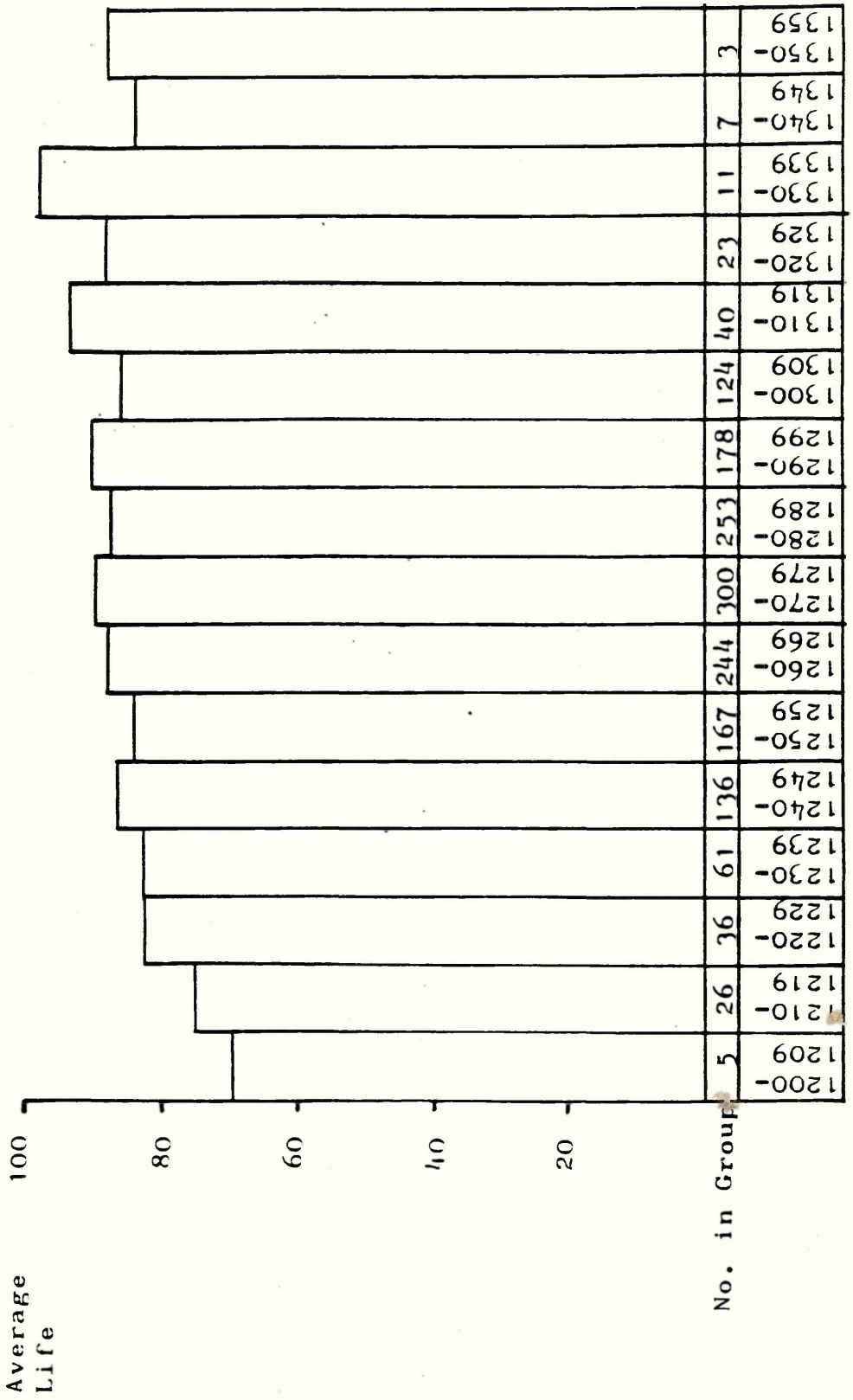


Figure 67 The Effect Of Casting Temperature On Performance Of 610WB Moulds

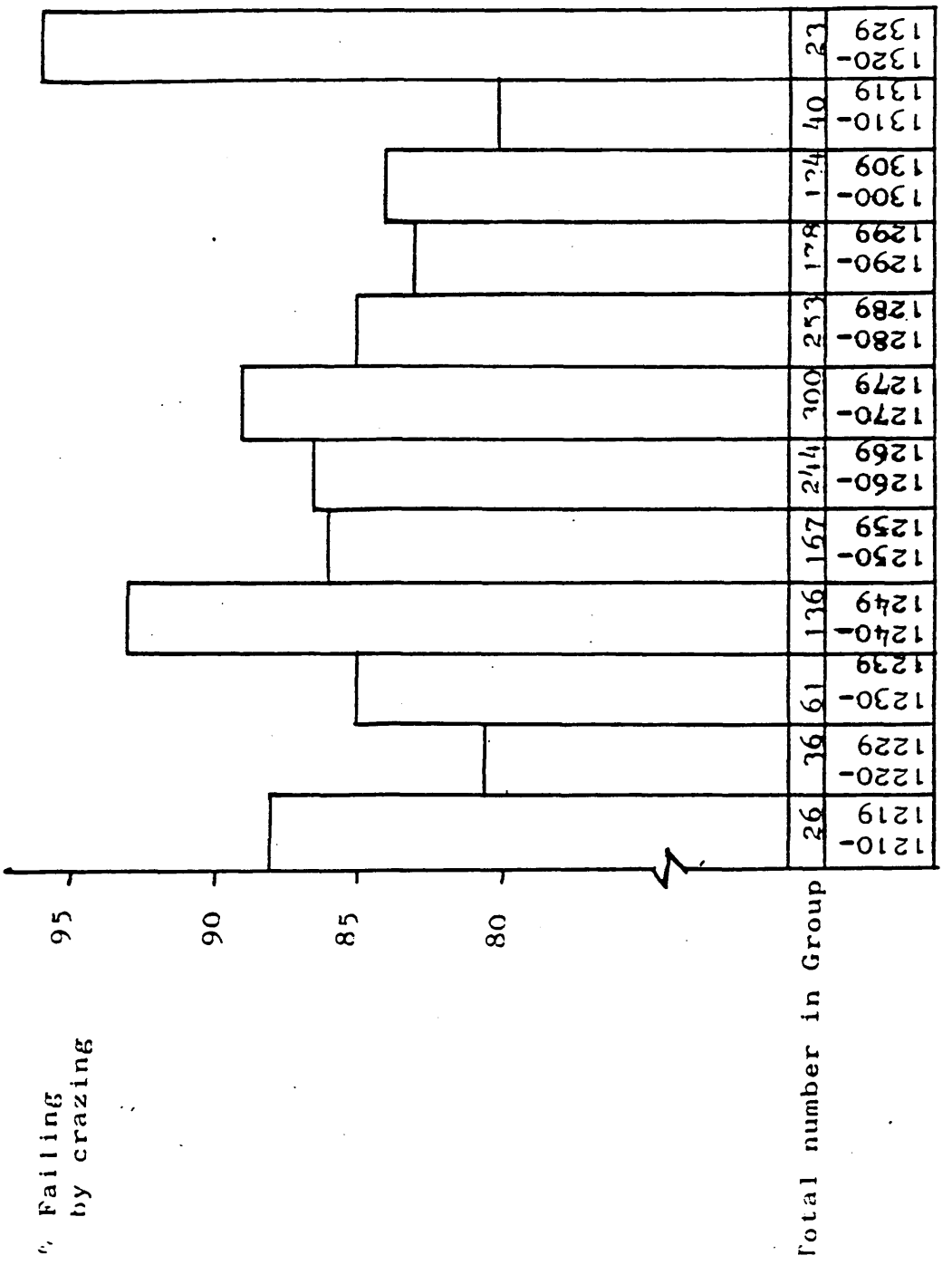


Figure 68 Effect Of Casting Temperature On Failure Mode Of 610WB Moulds

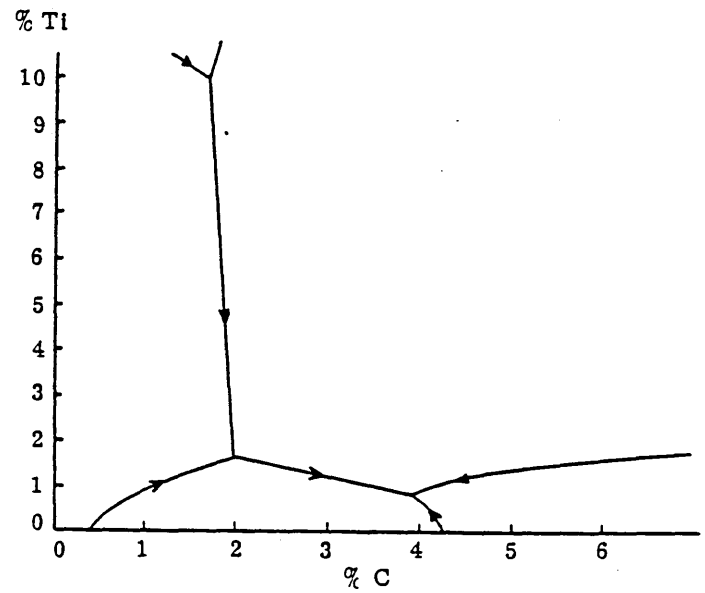


Figure 69 Liquidus Lines In The Iron-Rich Corner Of The Fe-C-Ti System

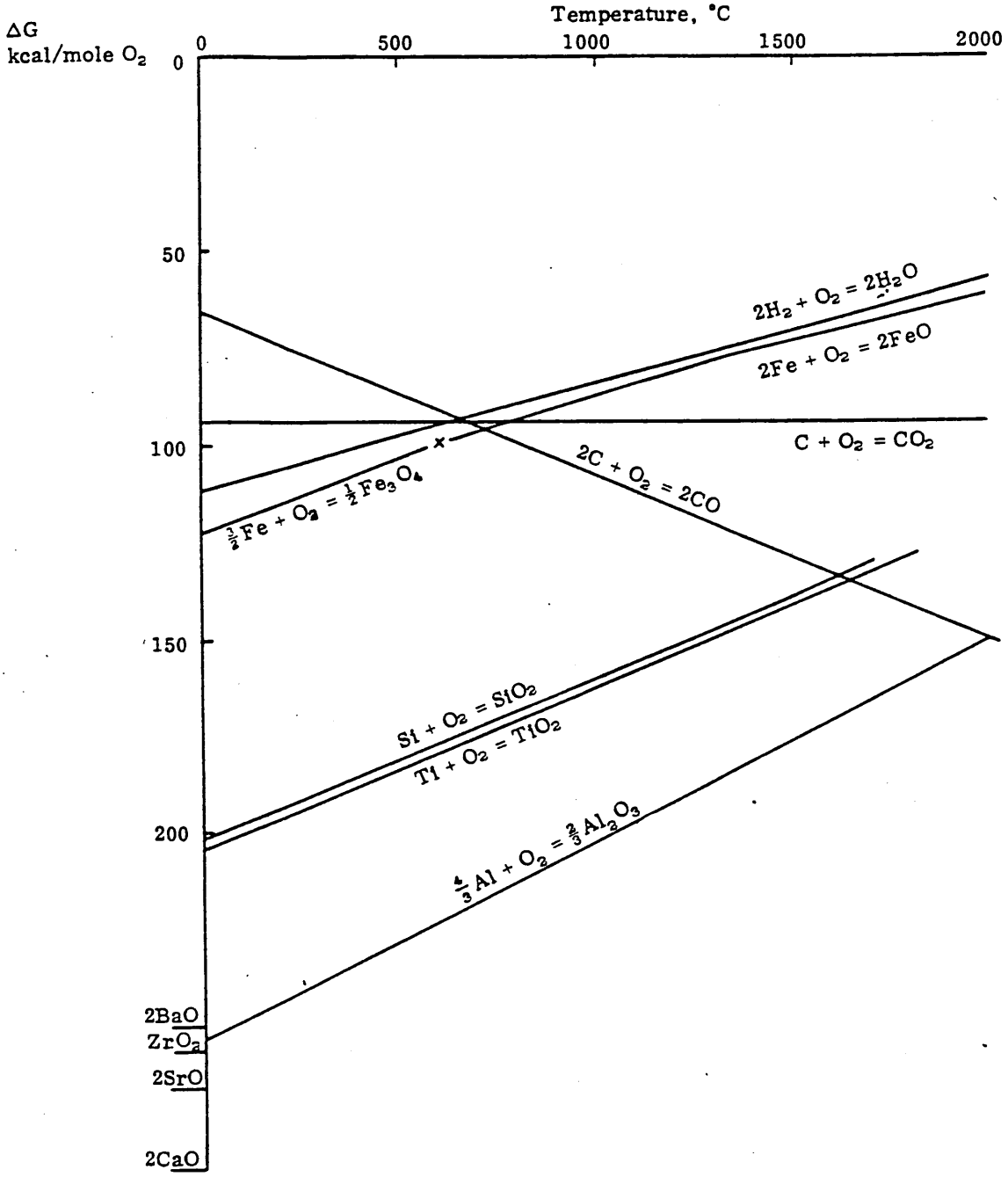


Figure 70 Free Energy Diagram For The Formation Of Oxides

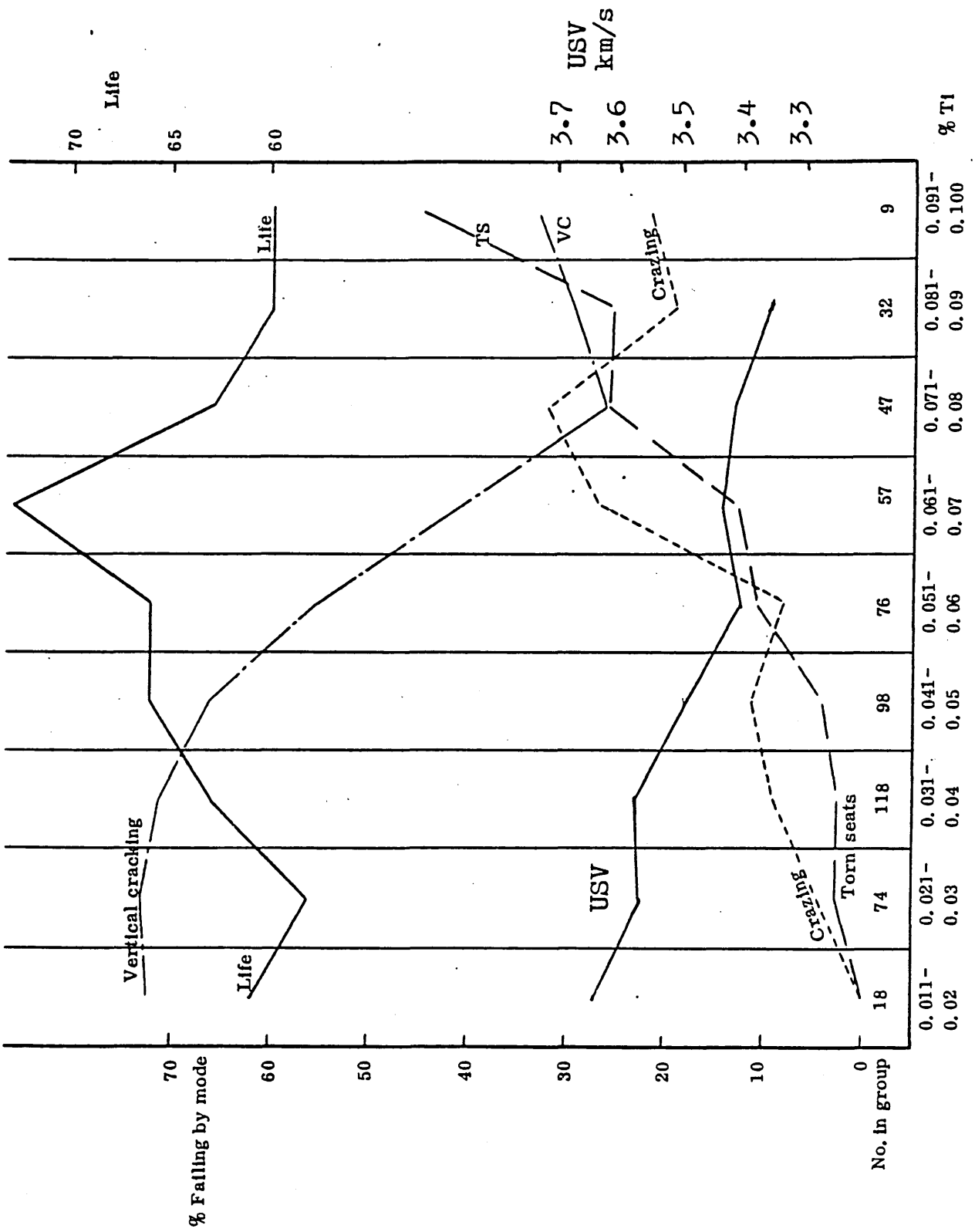


Figure 71 The Effect Of Titanium On Life, Failure Mode And Ultrasonic Velocity For 48-Type Moulds With Controlled Stripping

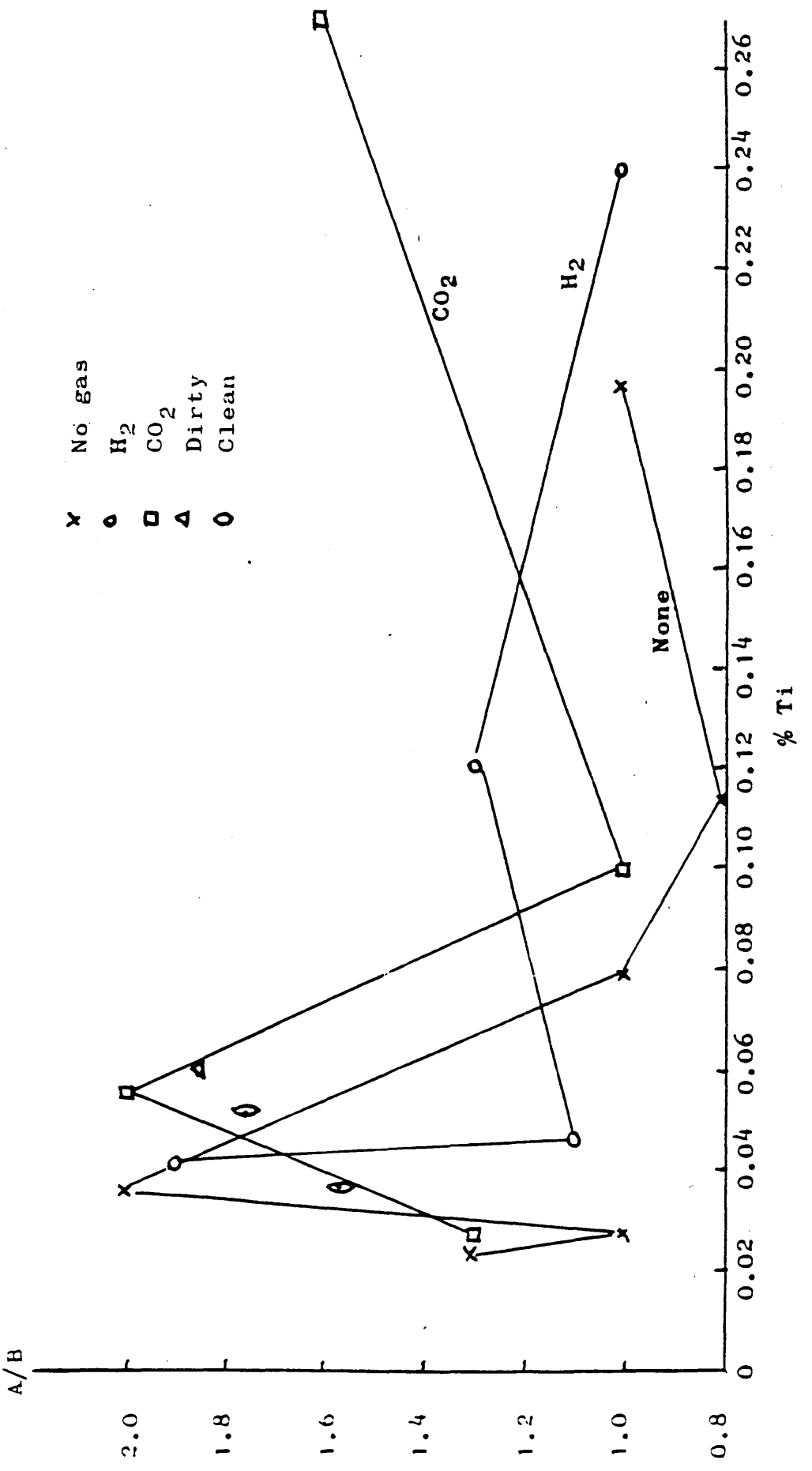


Figure 72 The Effect Of Titanium On The A/B Bend Test Parameter In Heavy Section Test Blocks

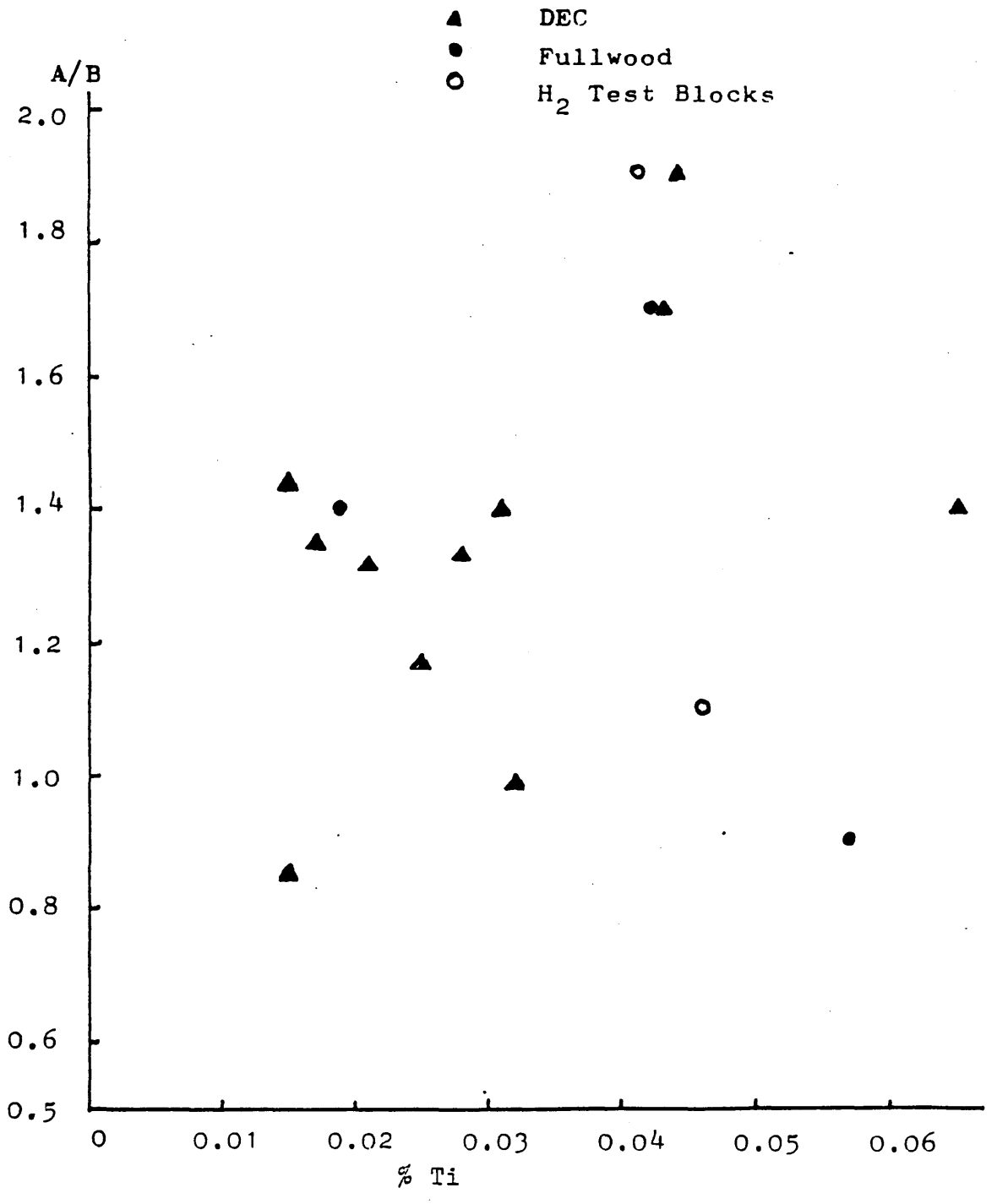


Figure 73 The Effect Of Titanium On A/B Bend Test Parameter In Cupola Melted Irons

DESCRIPTIONS OF B.S.C. FOUNDRIES WHICH HAVE OR ARE CURRENTLY PRODUCING INGOT MOULDS

1. Dowlais Foundry, FFE Group, Merthyr Tydfil

Dowlais are the most recent additon to BSC foundries, being part of G.K.N. until 1974. Dowlais melt in an acid, hot-blast cupola and use the dry-sand system for ingot moulds, which are produced in the 5 to 25t range and are, chiefly, of the slab type for the Port Talbot and Llanwern works of the Welsh Division.

Dowlais is the main producer of quasi-flake ingot moulds, almost the whole of their ingot mould product mix now being produced as quasi-flake iron.

2. Fullwood Foundry, FFE Group, Motherwell

Prior to 1974, Fullwood also used basic hot-blast cupolas as the melting furnaces. Since the early 1970's, however, a resin-bonded, furane sand system has been used. In 1979, the cupolas were removed and replaced by two 120t channel induction furnaces, designed to operate on a charge of whole unbroken ingot moulds returned from nearby steelworks after usage. Moulds are produced in the 4 to 40t weight range, principally being large slab moulds for the Ravenscraig steelworks of Scottish Division.

3. Renishaw Foundry, FFE Group, Sheffield

Renishaw use both divided blast cupolas (run to a cold blast acid practice) and 8t induction furnaces. Moulds are produced, principally, by desulphurising cupola metal in

a porous plug ladle, with calcium carbide and duplexing through the electric furnaces; direct cupola or electric furnace melting may be used as well. The sand practice used is the dry sand system.

The maximum mould weight produced is 8 tonnes, and are usually of the square section type used in the Sheffield Division.

4. Stanton, Stavely Foundry, Tubes Division, Stanton

Stanton Erewash foundry is part of the spun iron complex. The foundry takes its metal from the cupolas of the central melting plant and the iron is duplexed through a coreless induction furnace. The sand practice used is the fluid silicate system.

Moulds are produced in the 4 to 15t range and their principal customers were Corby and Bilston. With the closure of these plants, Stanton now supply, principally, to the Scunthorpe Works and to Sheffield Division, and have, latterly, tended to produce mainly SG iron moulds.

5. Distington Foundry, Cumbria Division, Workington (Closed June 1981)

Prior to its closure in June, 1981 Distington foundry was the largest ingot mould foundry within the Corporation. Iron used to be supplied by basic hot blast cupolas, and a dry sand system was employed. Moulds of up to 70t have been produced, the normal range was 4 to 25 tonnes.

In 1979/80, an extensive development was carried out, involving the replacement of the cupolas with 120t channel induction furnaces. These were designed to accept hot metal direct from the blast furnaces at Workington, and to trim the analysis by additions of steel scrap. Cold melting was also utilised, and the dry-sand system was replaced by a resin-bonded, furane system.

Distington's major customers were Teeside, Scunthorpe and Sheffield Divisions.

6. Landore Foundry, FFE Group, Swansea (Closed August,1980)

Landore's metal supply was from a hot blast acid cupola, the metal being trimmed and desulphurised in shaking ladles. The sand practice was the dry sand system.

Due to rationalisation of ingot mould production capacity, mould production virtually ceased in 1978, and the foundry closed in 1980.

7. Teeside Bridge, RDL, Middlesborough

Teeside Bridge foundry took its metal from an acid cold blast cupola, and the iron was continuously desulphurised with soda ash. The sand practice used was the dry-sand system.

Ingot mould production also ceased at Teeside Bridge, with the reorganisation in 1978.

8. RiverDon Works, FFE Group, Sheffield

River Don is primarily a steel foundry and forgemaster,

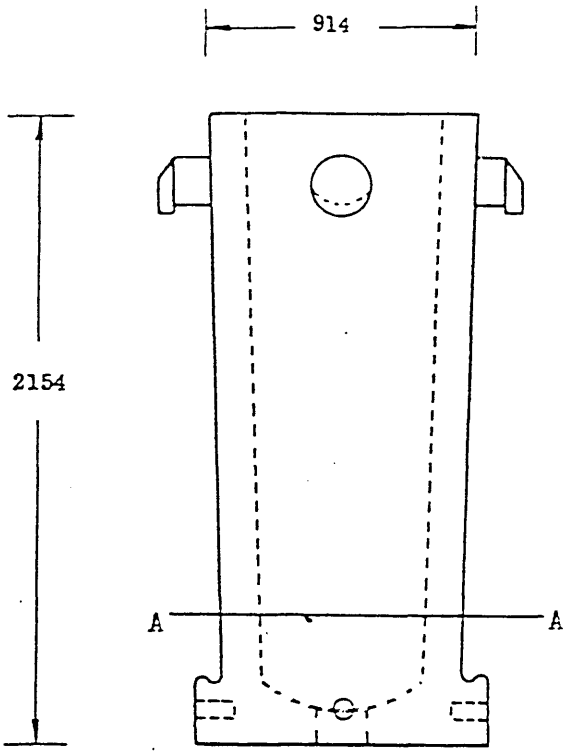
producing forgings up to 200t. The moulds for such forgings may weigh up to 150t, which is outside the capacity of BSC's ingot mould foundries. River Don Works have, thus, produced moulds of this weight for their own use.

The melting practice used is the basic electric arc furnace and trimming and any super heating is carried out in a Finkl Vacuum Arc Degassing Unit. The nominal furnace capacity is 100t, so moulds above this weight are produced in two parts, the first cast being held in the VAD unit whilst the second cast is melted.

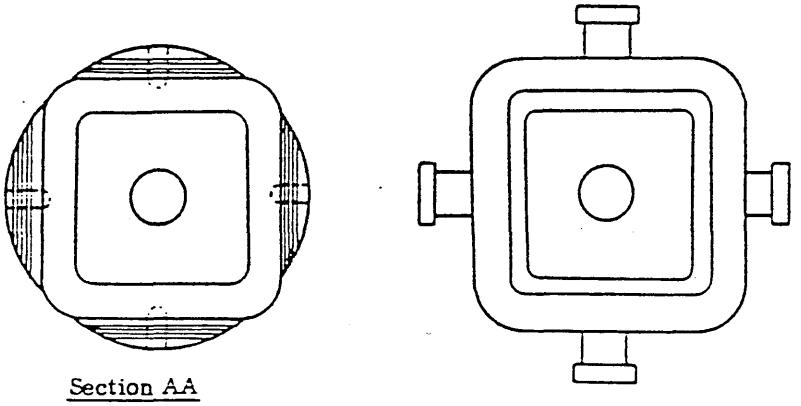
The sand practice is the resin-bonded furane system and the moulds are produced in large pits, as opposed to the normal boxed moulding practice.

9. Craigneuk Foundry, FFE Group, Motherwell

In 1978/79, Craigneuk steelworks, which incorporates a steel foundry, was short of ingot moulds (around 6t) and so briefly, produced their own. These were melted in a basic electric arc furnace of 25t capacity, the moulds being produced in the resin-bonded phenolic system.

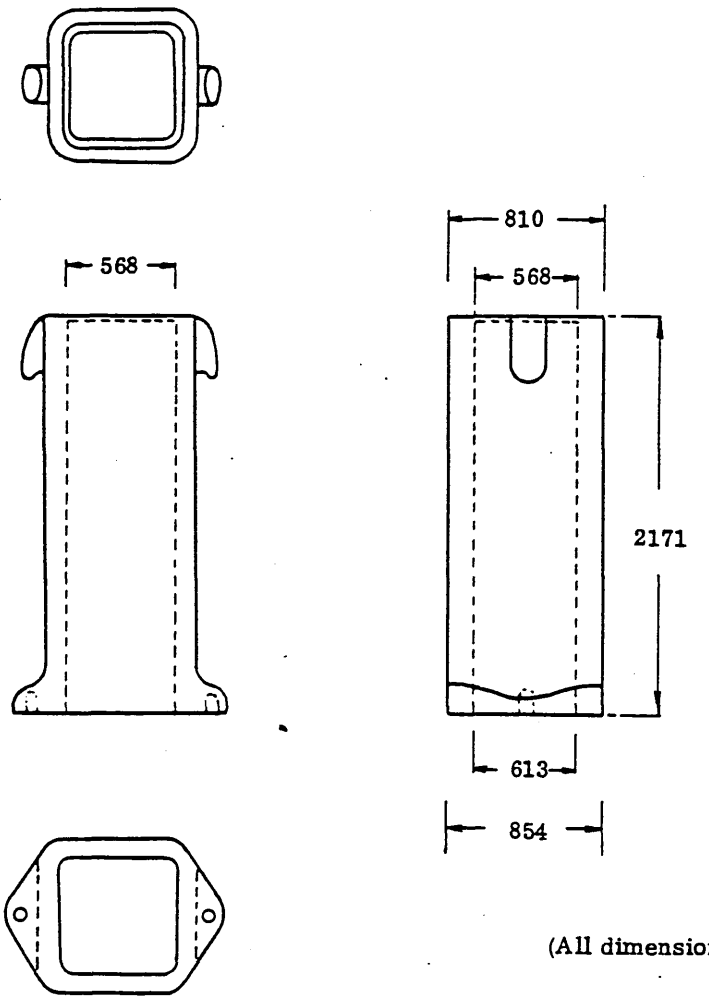


(All dimensions in mm)



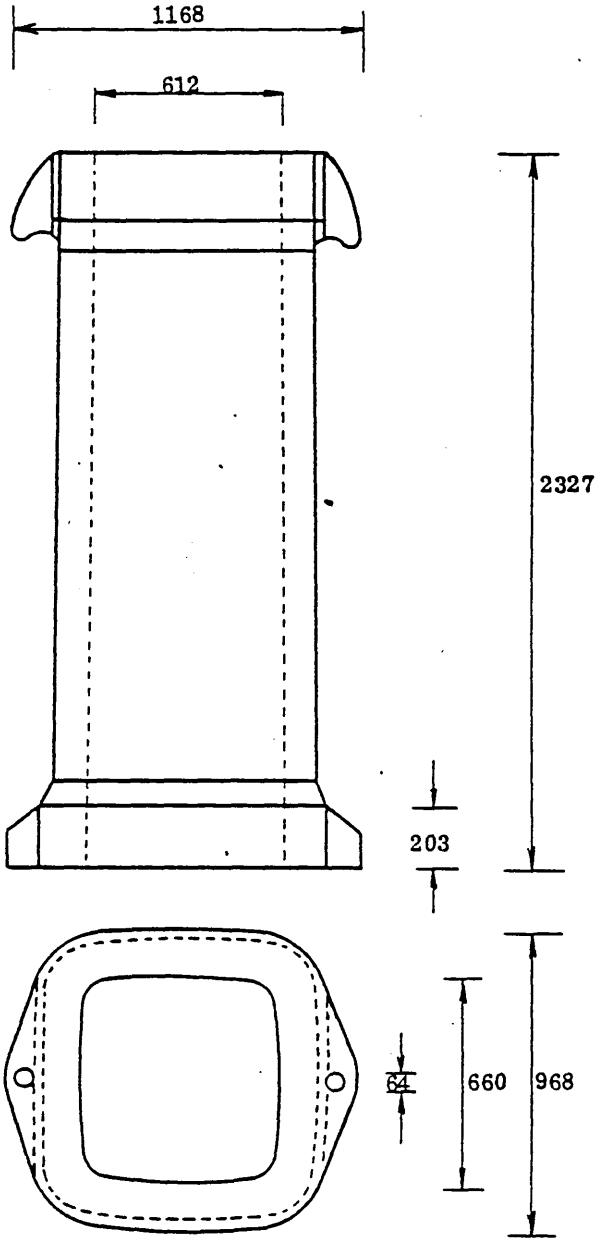
Section AA

A2/1 Tinsley Park F-Type Mould (6.1t)



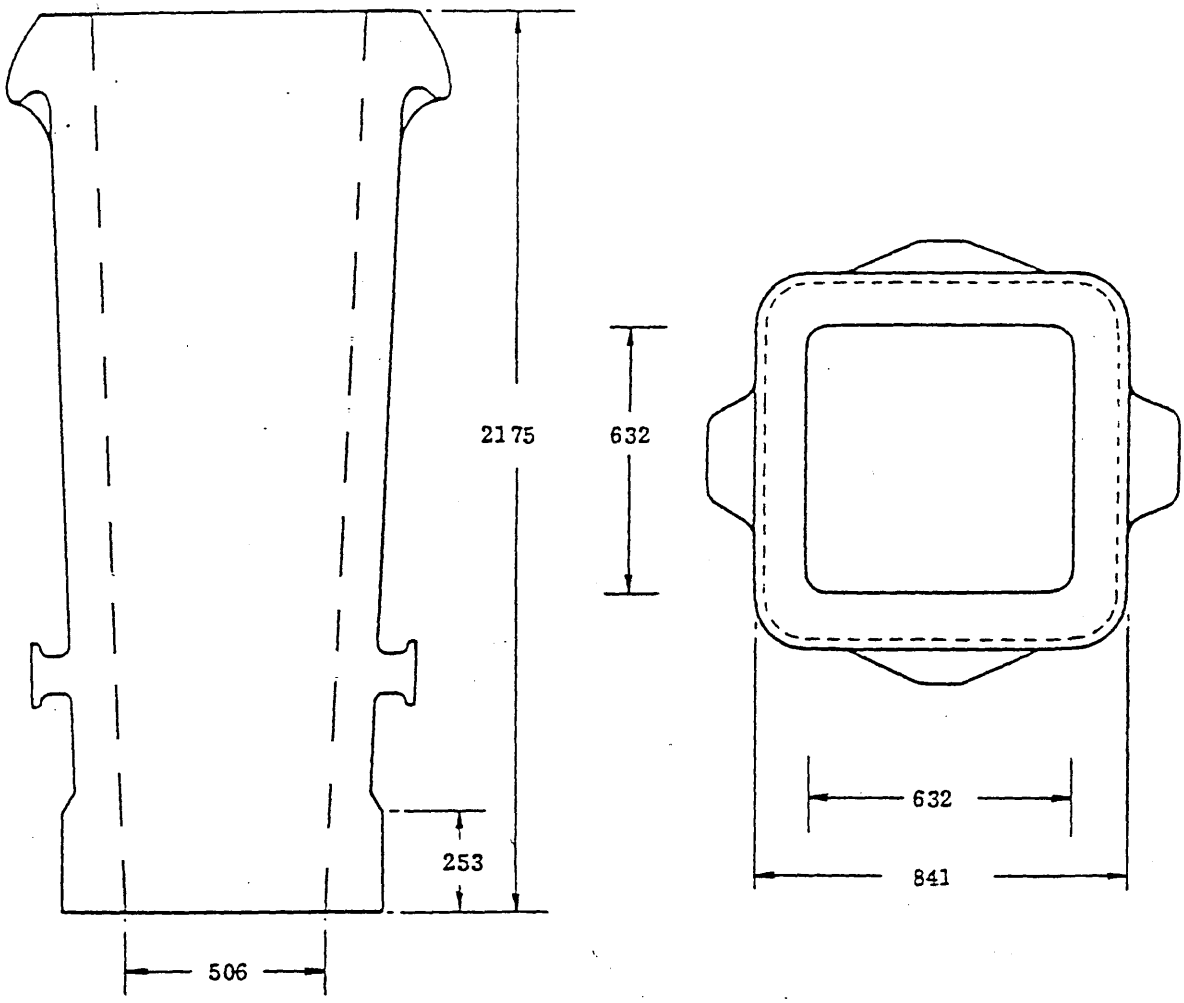
(All dimensions in mm)

A2/2 Templeborough 610 WB Mould (4.95t)



(All dimensions in mm)

A2/3 Aldwarke B120 Mould (6.4t)



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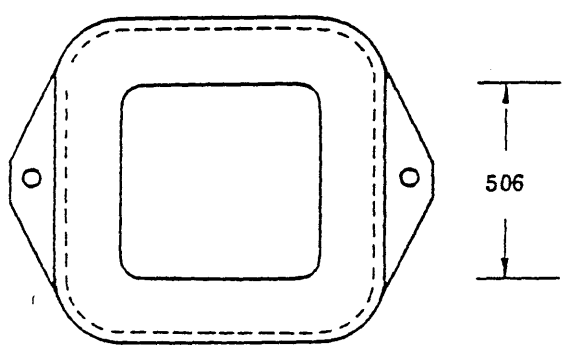
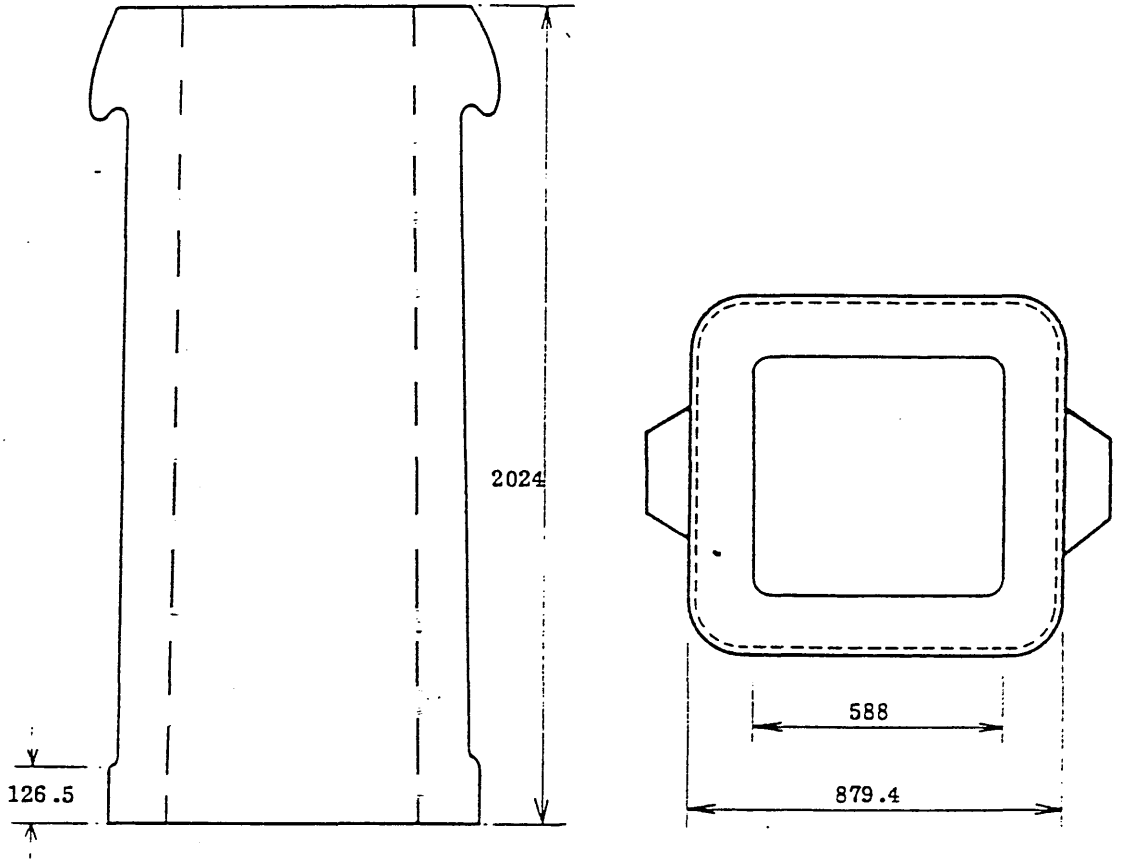
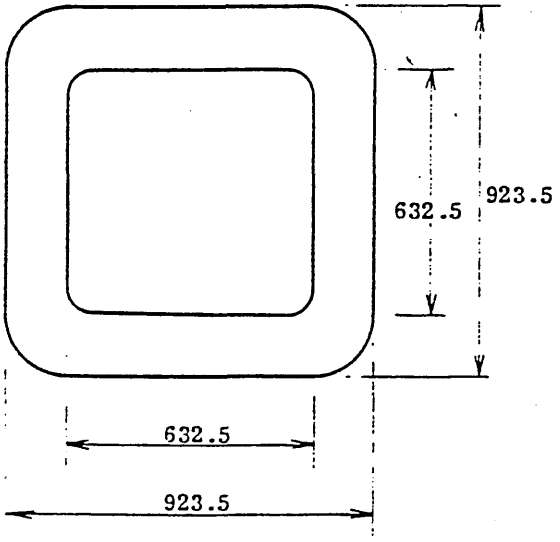


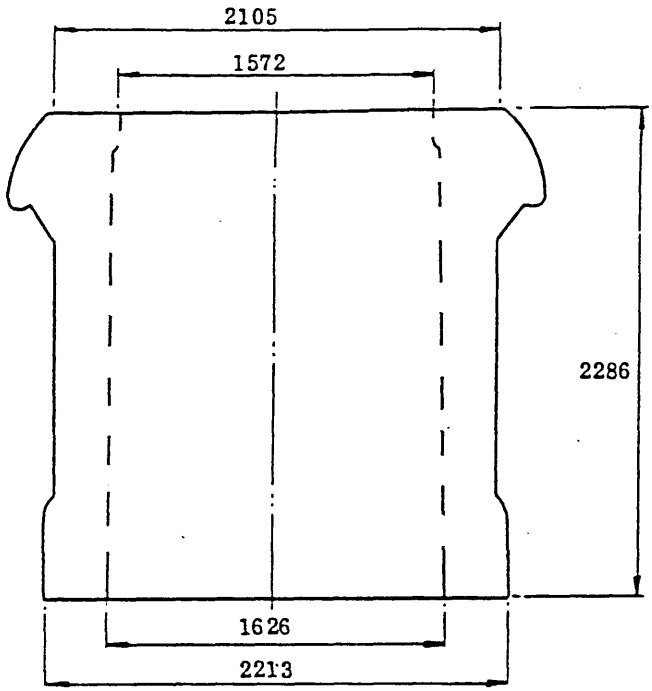
A2/4 Round Oak WEU 100 Mould (5.5t)

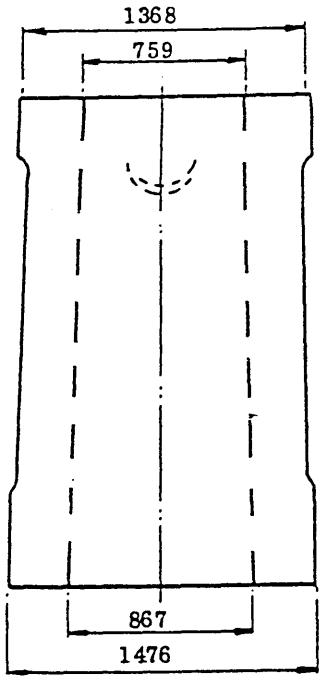


(All dimensions in mm)

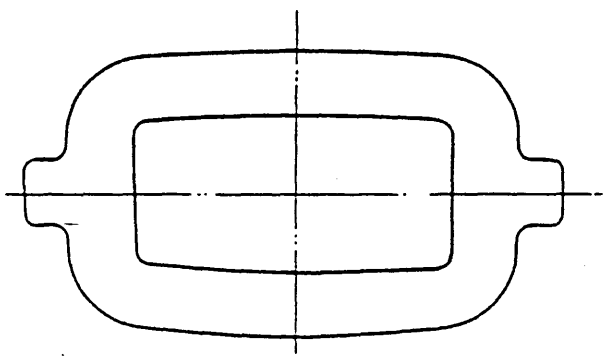


A2/5 Round Oak NEU 100 Mould (5.55t)

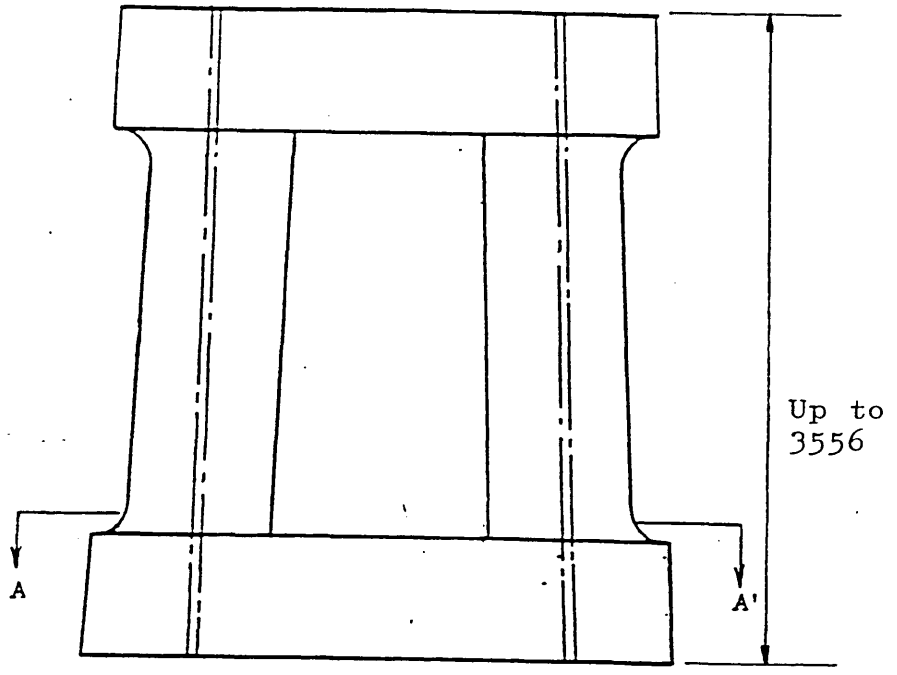


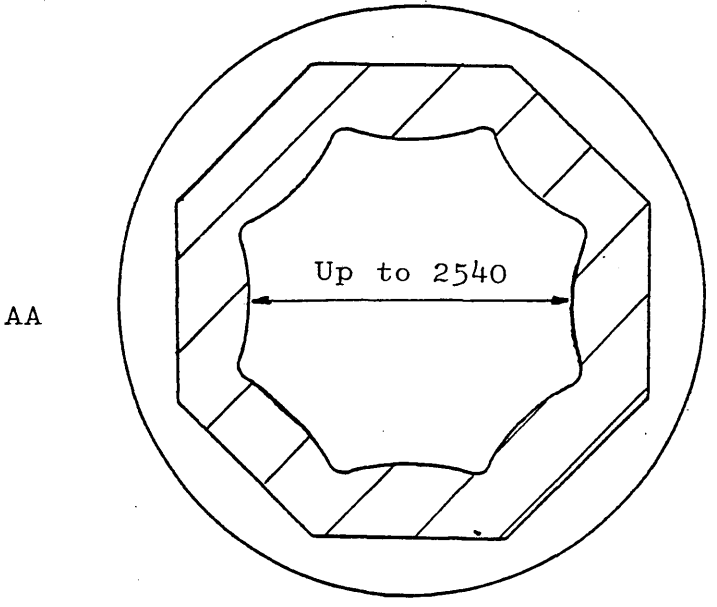


All dimensions in mm



A2/6 Ravenscraig 48-Type Mould (24.9t)





A2/7 Typical River Don Forging Mould

CODE OF PRACTICE FOR HEMATITE IRON INGOT MOULD MANUFACTURE

ISSUED BY THE IRON CASTINGS TECHNICAL LIASON COMMITTEE

8th March 1976

Ingot mould performance is governed by foundry (manufat-uring) and steelworks usage parameters. Within the last four years, Corporation steelworks have established a code of operating practice, designed to optimise ingot mould performance.

In order to gain benefit from the steelworks' code of practice, it is important that ingot moulds supplied by the foundries be of the correct chemical analysis and metallurgical structure, be mechanically sound and free of surface defects.

This document, therefore, summarises the main foundry parameters, which it is necessary to control if these characteristics are to be acheived. Thus, physical properties are determined by the metallurgical structure, which, in turn, is dictated by the stripping practice (Section 7) and chemical analysis (Section 8). The soundness and integrity of ingot moulds is affected by sand (Section 1) and running/casting practice (Section 3-6). The best ingot mould product may be achieved only by attention to the parameters identified in this document.

1. FOUNDRY SAND PRACTICE.

Ingot moulds within the U.K. are produced, mainly, by the dry

sand process, although other systems (furane-resin bonded and fluid sand) are employed, currently.

There are, inevitably, minor variations between the foundries utilising the dry sand practices and these relate to the ammount of return sand used, different additions and the methods of compaction (sand slinging or jolting). Both the furane and fluid sand systems rely on chemically hardening the sand moulds, both systems being examples of 'cold set' techniques.

Irrespective of the individual sand practices employed, the aim of the moulding practice is to produce a rigid, permeable sand mould with adequate high temperature properties, in order to produce a satisfactory final product.

Any sand process which produces a sound, dimensionally accurate ingot mould, free of surface defects, is acceptable.

2. INGOT MOULD DIMENSIONS

Ingot moulds produced in the Corporation foundries should comply with the following dimensions;

Mould Wt. Range	Wall Thickness Tolerance	Ingot Cavity Tolerance*	Resulting Ingot Wt. Tolerance*
10 tonne	± 3%	± 0.7%	± 1.5%
10-20 tonne	± 5.5%	± 2%	± 1.5%
20 tonne	± 5.5%	± 2%	± 1.5%

* These tolerances do not imply a maximum tolerance for broad and narrow cavity dimensions simultaneously. For

this reason, the spread in anticipated ingot weights ($\pm$ 1.5%) represents a better reflection of the spread in mould cavity dimensions.

There is no evidence to suggest that variations in wall thickness of the magnitude indicated impair mould performance provided that mould symmetry is obtained.

The location of plug holes, lifting lugs or feeder head slots should comply with refractory and steelworks tolerance requirements.

3. CASTING TEMPERATURE *

In order to minimise run-out, sand burn-on and shrinkage as a result of liquid shrinkage the casting temperature should be as low as possible consistent with preventing cold shut/lapping problems. On the other hand, casting temperature has an effect on the mechanical properties (and hence performance) of ingot mould iron, optimum properties being acheived following casting from 1250 - 1280°C. The actual casting temperature is therefore a compromise, designed to produce the best product. The temperature employed is usually a function of the section size, and for ingot moulds, casting temperatures in the range 1250 - 1280°C are employed.

* Casting temperature is defined as the temperature of the hot metal as it enters the sand mould. If the temperatures are measured at other points in the production sequence,

due allowance should be made to ensure that casting temperatures fall in the defined range.

4. POURING RATES

Pouring of iron should be smooth and uninterrupted. In dry sand moulds, excessive radiant heat in the mould cavity during filling can lead to "scabbing" on the casting surface. In resin bonded sand moulds, radiant heat can lead to the breakdown of the mould due to burn-out of the resin. Mould filling should therefore be carried out as quickly as possible taking account of practical constraints. Rates of 3 tonne/minute for small (<5t) moulds and 6t/minute for larger moulds are recommended. The exception is large (>60t) octagonal moulds which should be poured at 20t/minute.

5. RUNNING SYSTEMS

Running systems must accept iron at the appropriate pouring rate (see 4) and should be designed to deliver iron as smoothly and uniformly as possible in order to minimise erosion. The ingate configuration is practically determined by moulding requirements, but as far as possible the design should avoid direct impingement of hot metal streams onto core faces or corners. Any system which achieves these requirements is satisfactory.

6. FEEDING

The aim of ingot mould foundry practice should be to produce a sound, dimensionally accurate product. The solidification of grey cast iron (~4%C) is accompanied by a slight increase in volume, due to the precipitation of graphite. In the case of ingot moulds, particularly of the larger (>15t) types, casting of the iron into the mould may, however, be accompanied by mould dilation, due, probably, to the ferrostatic pressure and the expansion characteristics of the iron during solidification. The extent of the dilation and liquid shrinkage is a function of iron analysis, metal temperature and mould size, but the amount of feed metal required may be as high as 4% of the mould weight.

Feeding practice in ingot mould foundries should, therefore, use feeder heads of sufficient size to produce moulds free of shrinkage, both at the mould top surface and within the body of the mould. The size and disposition of the feeder heads is likely to vary from foundry to foundry, but the maximum size of feeder heads is unlikely to exceed 4%. In the case of the large feeder heads, the use of Washburn cores will facilitate the easy removal of the heads, which may then be remelted. Under no circumstances should feeder heads be removed from ingot moulds by burning.

7. STRIPPING PRACTICE.

The effect of the rate of cooling of ingot mould and bottom plate castings during manufacture upon the structures deleloped within the iron and, in consequence, upon the service

behaviour of the castings in steelplant service has been demonstrated to be important.

In the manufacturing process, the factor which controls the cooling rate is the time interval after pouring for which the casting remains completely within its sand environment. Removal of the casting from the sand to atmospheric cooling should occur when the casting has cooled under such control to a temperature of the order 700°C.

The thickness of the sand compact covering the whole surface of the casting should be sufficient to provide slow cooling. In the case of dry sand, on which most experience has been gained, this corresponds to a thickness of 125-150mm. In the case of furane bonded sands, recently acquired information suggests that similar cooling rates can be achieved with 75-100mm of sand.

Greater thicknesses of sand have no significant influence on the cooling process. Inadequate sand thicknesses, particularly if backed by the metal of the casting tackle, even if they occur only at localised positions on the casting can have a significant influence on the cooling rate of a major part of the casting.

The extremities of ingot mould walls require particular care in the cooling operation since they become prone to exposure to atmospheric cooling during the various stages of breakdown of the moulding tackle within the manufacturing process route.

During the control cooling period, it is necessary only that the moulding sand compact should remain in intimate contact with the ingot mould casting body. Whether or not it is necessary to retain the metal moulding tackle around the sand to acheive this situation will depend upon the type and characteristics of the moulding sand, upon the handling, etc., methods employed within a particular manufacturing route.

Bottom plates should be cast within a mould having a minimum sand thickness of the order of 150mm (or within a refractory mould having equivalent heat insulation properties.). The upper surface of the plate casting should, preferably, be covered with a top part carrying a thickness of sand compact or other refractory material with heat abstracion properties equivalent to those of a minimum thickness of the order of 125-150mm of dry sand compact. The top part, or at least its sand/refractory material content, should remain in position above the casting until the latter attains a temperature of the order of 700°C.

Prediction of the cooling times for which it is necessary that ingot mould castings should remain within the sand mould, in order that the mould wall temperature should be of the order of 700°C, may be made from Fig. 1. The cooling time interval 'pour-strip' in hours is related to the ratio of the mould weight in tonnes to the total surface area of the inner and outer walls of the ingot mould in square metres.

Prediction of the necessary cooling times for which bottom plate castings should remain totally within the sand mould in order that the plate surface temperature shall be of the order of 700°C is more difficult to define in a generalised form. The individual plate geometry has a significant effect upon the relationship. An approximation to the time interval 'pour-remove top part' in hours can be obtained by multiplying the ratio of the volume of the plate in cubic metres to the total surface area of the plate in square metres by 250. The volume and surface area value used can be computed from the respective overall average dimensions of the plate. Refined values for the cooling times for individual plate designs to be applicable within control procedures can be determined on the initial approximations.

8. TARGET ANALYSIS LEVELS

Proposed limits for an analysis range for ingot moulds must assume that the cooling conditions of the mould and the stripping practice complies with that specified in (7).

The five major elemants for normal flake graphite ingot moulds should fall within the following range:-

C	Si	S	P	Mn
3.7-4.0	1-2	0.1 max	0.15 max	0.5/1.1

There are certain cases where varaitions outside this range, particularly carbon content, have produced satisfactory performaces in moulds.

The wide range of silicon reflects variations in mould

requirements. Silicon is a graphitising element and the lower silicon contents may be used where pearlite structures, with a good resistance to crazing, are required. Higher silicon levels are used to increase the amount of ferrite in moulds where the normal mode of failure is by cracking.

Certain trace elements may affect the structure of ingot moulds, particularly if the controlled cooling practice (7) is not maintained. Thus, the presence of strong, carbide stabilisers, notably chromium, is known to be detrimental. In addition, it has been established that in small (7t) moulds, increasing quantities of copper (pearlite stabiliser) impair mould performance and examinations of individual mould failures have shown that the presence of excessive amounts of tin (0.07%) and lead (0.006%) are also associated with premature mould failures. Compacted forms of graphite appear to be promoted by high nitrogen contents, but if they are undesirable they may be prevented by addition of aluminium or titanium.

The precise levels at which trace elements have an effect on mould microstructure have not been established, although information currently being accumulated will provide guidance on acceptable levels. The situation is further complicated by the fact that the effects of trace elements is cumulative. The subject has been considered in some detail by the British Cast Iron Research Association (Document IMP 138 'Trace Elements in Cast Iron and Their Possible Significance in Ingot Moulds'). On the basis of current experience and published data, the following analyses are

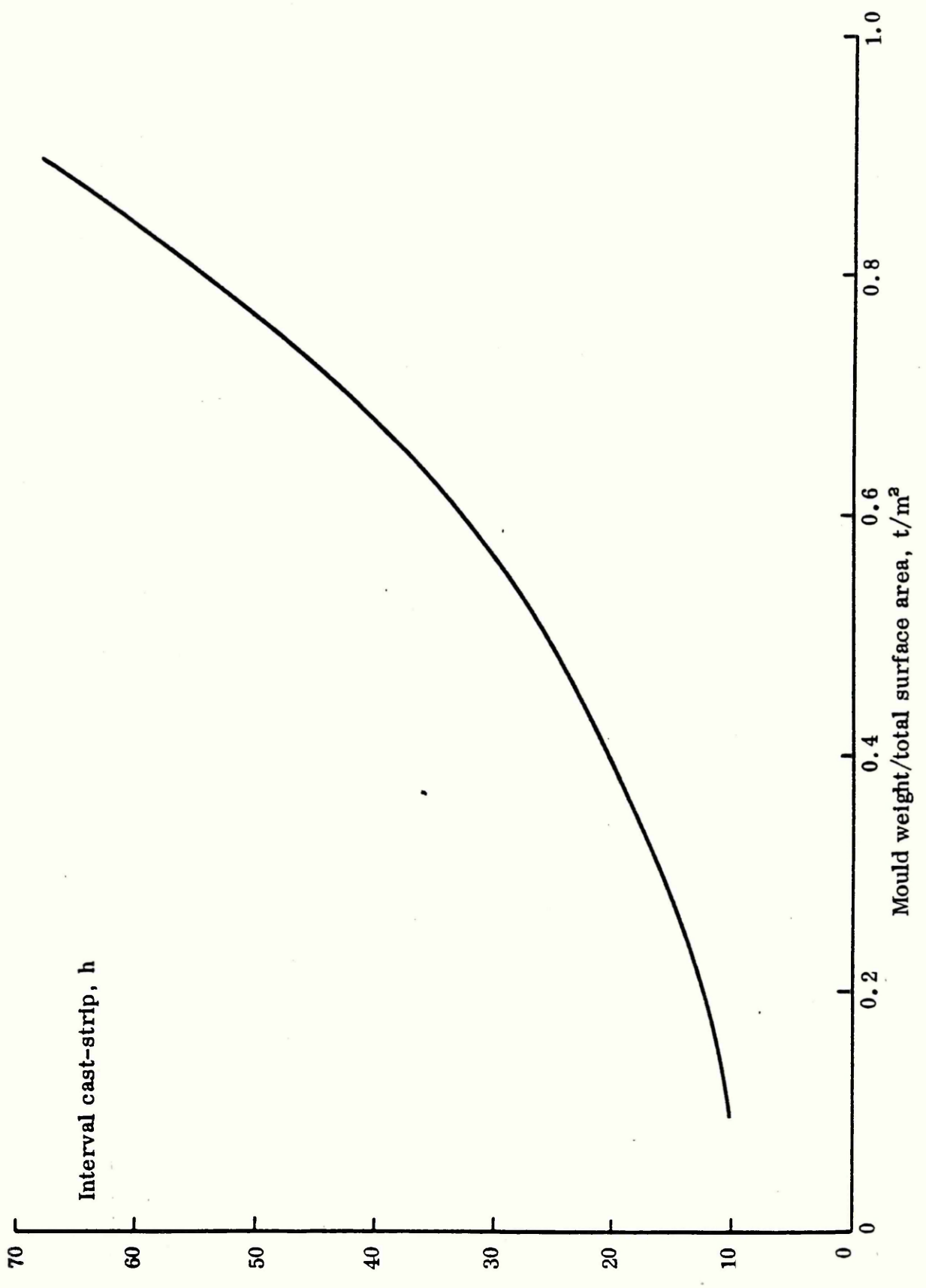
suggested as target levels in ingot mould manufacture. The levels are likely to require modification as more information becomes available:

Pearlite Stabilisers				Carbide Stabilisers	
%				%	
Sn	Cu	Ni	As	Cr	Mo
0.02 max	0.3 max	0.2 max	0.01 max	0.05 max	0.01 max

Others.

%	
N	Pb
0.007 max.	0.0005 max.

It must be emphasised that a large number of moulds in current service exceed these individual levels without any apparent deterioration in mould performance.



A3/1 Cooling Of Ingot Mould Wall To 700°C Prior To Stripping

THE EFFECTS OF PHOSPHORUS AND FOUNDRY STRIPPING PRACTICE ON THE PERFORMANCE OF CAST IRON INGOT MOULDS

K.B. Wilford and F.G. Wilson

BRITISH STEEL CORPORATION, SHEFFIELD LABORATORIES

Phosphorus levels in haematite ingot moulds are normally maintained at 0.12% max. There is evidence, however, to show that levels up to 0.23% may be beneficial. Due to a shortage of suitable raw materials, a BSC ingot mould foundry operated for a period of 6 months with a high phosphorus cupola burden and levels up to 0.264% P were produced. Of the mould types produced during this period there was a slight drop in performance of up to 7 lives for NEU open topped moulds but the performance of a WEU closed bottom mould deteriorated from 120 lives down to 70 lives due to cracking. Experimental casts showed that high phosphorus levels caused an increase in brittleness, this effect being enhanced by high residual levels and by fast cooling from 1000 °C. Microstructurally this was associated with the formation of cell boundary networks of phosphide/carbide eutectic. Surface temperature measurements on actual moulds showed that on removing the top plate the temperature of the WEU CB mould could be as high as 960 °C compared with less than 750 °C for the NEU OT moulds. This is due to the heavy base section (top as-cast) of the closed bottom mould. The poor performance of the WEU CB mould has therefore been attributed to high phosphorus levels (above 0.12%) coupled with fast cooling rates.

1. INTRODUCTION

British Steel Corporation currently produces about 200 000 t/a of cast iron ingot moulds in the iron foundries for use mainly within the Corporation's steelworks. At the present time the annual value of the moulds, which generally fall within the weight range 3-30 t, is £40-M. The steelworks' performance of these moulds therefore has a considerable bearing on the operating costs of the Corporation's steelworks.

Within the past few years, considerable success has been achieved by the production and use of modified graphite (spheroidal and compacted) ingot moulds and the use of such moulds is continuing to increase. However, the bulk of moulds

used within the Corporation is still of the conventional flaké graphite type and considerable attention continues to be paid to the performance of haematite moulds.

An ingot mould may be defined as a container into which liquid steel is poured for the purpose of producing a solid ingot for subsequent rolling or forging. Practically the design of the ingot shape is based on solidification and yield considerations and the mould is designed to contain the steel and conduct heat away whilst it solidifies. During usage, therefore, the inside surface of the mould rapidly heats up to temperatures of the order of 700-900 °C followed by equalisation of the temperature gradients within 5-15 minutes. The moulds after use are allowed to cool below 100 °C and reused.

The number of times the moulds can be reused before they require scrapping is an important factor in the economics of the ingot route. During usage the inside surface deteriorates through thermal fatigue, growth and oxidation. Under optimum conditions, therefore, the deterioration in surface quality means that ultimately the ingot cannot be stripped, or the resultant billet or slab surface quality is so poor that the increased dressing costs would not justify continuing use of the mould and so it is scrapped.

Frequently, however, moulds may fail before this condition is achieved. The most common cause for rejection at an early life is because of cracking.

2. FACTORS WHICH DETERMINE INGOT MOULD PERFORMANCE

Ingot mould performance is dictated by mould design, steelworks usage conditions and foundry manufacturing conditions. It is often difficult and sometimes impossible from the available data to decide which of the above factors contributes to poor mould performance, particularly in the case of individual premature mould failures.

This paper deals with a particular experience in which poor mould performances could be attributed directly to manufacturing deficiencies. Before describing the problem, it is considered appropriate to outline briefly the main foundry parameters which dictate mould performance.

3. FOUNDRY PARAMETERS

The foundry parameters which determine the performance of a given type of ingot mould are the properties of the iron, the physical soundness and dimensional accuracy of the mould. Thus, the iron needs to withstand both physical and thermal shock, the ingot produced has to be of good surface quality and dimensionally acceptable, while the dimensions of the mould have to be compatible with steelworks' requirements. All of these factors are determined by conditions in the foundry, where lack of control can lead to a deterioration in mould performance.

The present report deals specifically with one aspect of the properties of ingot mould iron (i.e. its cracking resistance) and factors controlling the soundness and dimensional accuracy are

not considered. Clearly the properties of the iron are determined by its microstructure, which in turn is determined by the melting route, the chemical analysis, metal treatment techniques (e.g. inoculation) and finally, the cooling rate of the casting.

The two features highlighted in the present paper are chemical analysis and cooling rate of the casting.

3.1 Iron Analysis

As a general aim, Corporation ingot moulds, which are produced via cupola or electric furnace routes, are manufactured to the following analysis requirement:-

C	Si	S	P	Mn
3.7-4.0	1-2	0.1 max	0.15 max	0.5/1.1

The properties of the cast iron, and consequently mould performance, is determined partly by chemical analysis. For example it is well established that increases in carbon equivalent reduce the tendency for cracking to occur but also accelerate the crazing process. Analysis variations can therefore be made within the quoted range to allow for differences in mould size and design and steelworks' operating conditions.

Although it is acknowledged that residual elements affect iron properties and mould performance, statistical analysis of performance data is inconclusive in this respect. Every effort is made to restrict known subversive elements (e.g. lead) to acceptable levels, but in general, residual levels are dictated by the use of normal raw materials.

The phosphorus levels in ingot moulds produced in the Corporation foundries are normally maintained in the region 0.04-0.08%. However, the third report of the Ingot Mould Sub-Committee[1] provided evidence to show that phosphorus contents in excess of those normally associated with ingot mould iron may be beneficial for the performance of small moulds (4 t). The range considered was 0.06-0.23% P.

3.2 Cooling Conditions

Ingot moulds in BSC foundries are produced using different sand systems (e.g. dry sand, silicate, furane) but the manufacturing route employed is essentially similar, irrespective of sand system and moulding method employed. Most Corporation moulds are produced in boxes which employ sand thicknesses in excess of 150 mm.

It is an established fact, based on experience over the past 20 years, that premature stripping of the tackle in the foundry (i.e. removal of the box and exposure of the ingot mould above 700 °C) can produce a material which is susceptible to cracking. This is reflected in a deterioration in mould performance, usually because of premature cracking.

4. ANALYSIS OF INGOT MOULD DATA

During 1974 the quality of steel scrap available to one of BSC's foundries (designated Foundry A) deteriorated, both in size and trace element levels. A supply of railway chairs (40 kg each with 0.75-1% P and low residual content) became available and so these were charged at a nominal rate of 10-15% to the cupola. The mean phosphorus level resulting from this burden should have been approximately 0.16% which is slightly higher than the normal maximum allowed level but within the range mentioned above.

From July to December 1974, all moulds produced at Foundry A were produced from the new cupola burden. A variety of mould types was produced in the 4-7 t weight range, these principally being of the narrow end up (NEU) open topped (OT) and wide end up (WEU) closed bottom (CB) types. In particular, WEU, CB moulds with a heavy base were supplied to Steelworks S.

During December 1974, a number of premature failures mould occurred at Steelworks S, due to major cracking initiating from the base, the failure being mainly confined to high phosphorus moulds. Concern was expressed not only at the increased steel costs but also at the potentially serious safety hazard associated with the catastrophic cracking of a mould during teeming.

It should be emphasised that similar performance problems were not experienced at Foundry A's other ingot mould customers at the time. However, the situation was so serious at Steelworks S that as an immediate action, the phosphorus in the burden was reduced to its previous level by eliminating the railway chair scrap.

In order to clarify the effects of phosphorus on ingot mould performance a statistical analysis of the various mould types produced with high phosphorus levels was carried out. In addition a series of laboratory melts was made to study the effect of high and low residual element levels and phosphorus variations and foundry cooling conditions on the mechanical properties of the iron.

4.1 WEU CB Moulds at Steelworks S

Mean chemical analyses of moulds over a 2 year period ending with the 6 months of high phosphorus operation were as follows:-

C	Si	Mn	S	P	Mo	Cr	Sn	V	Ti
3.79	1.23	0.93	0.080	0.063	0.016	0.080	0.006	0.016	0.018

Ni	Al	Cu
0.057	0.002	0.106

Monthly average phosphorus contents and performances are shown in Fig. 1. Considerable scatter in the phosphorus levels occurred with values ranging from 0.035% up to 0.264%. It can be seen from Fig. 1 that from August 1974 when the phosphorus level was increased to a mean of over 0.11%, there was a large

decrease in performance from 115-125 lives down to 60-90 lives. Analysis of all moulds over the 2 year period showed that where phosphorus contents were below 0.13% the mean life was 120.2 (σ = 26.5) whereas moulds with over 0.131% P gave a mean life of 68.5 (σ = 26.3).

Multiple regression analysis for all moulds in the 2 year period was carried out and the following equation obtained:-

$$\text{Life} = 129.6 - 440.2\ (\%\ P) - 138.9\ (\%\ Cr) + 426.4\ (\%\ V) + 326.8\ (\%\ Ni) \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots \quad (1)$$

95% confidence limits, $2\sigma = \pm 56$

this equation being valid within the analysis limits given below:-

P	0.026 - 0.264%
Cr	0.001 - 0.39%
V	0.001 - 0.105%
Ni	0.001 - 0.192%

4.2 NEU OT Moulds at Steelworks T, U and V

Similar statistical analyses were carried out for NEU OT moulds at Foundry A's other 3 principal customers. No such marked trend was observed and high and low phosphorus content moulds had the following lives:-

Steelworks	Mould Wt. t	Performance	
		<0.13% P	≥0.131% P
T	4.95	84.2	77.4
U	6.44	106.4	103.2
V	5.55	95.1	109.9

It can be seen that at Steelworks T and U there has been a slight reduction in performance and at Steelworks V an apparent increase, although there are only 8 moulds of this type so that the results are not statistically significant.

5. EXPERIMENTAL CASTING PRODUCTION

It is apparent from the results in Section 4 that the premature failures of the WEU CB moulds at Steelworks S and to a lesser extent in those moulds at Steelworks T and U, were due to phosphorus contents in excess of 0.13%. At S, the average mould performance dropped from approximately 120 lives with low phosphorus (< 0.07%) to 68 lives for high phosphorus (>0.13% P) moulds.

The NEU moulds at Steelworks T, U and V, however, showed little apparent decrease in life at high phosphorus levels.

In an attempt to clarify the reasons for these inconsistencies, a number of laboratory melts were produced to study the effect of microstructural variables on cracking resistance.

250 kg melts were produced in an induction furnace using a charge of 100% Bremanger pig iron or 50% Bremanger and 50% Warner pig irons, to simulate different residual levels (designated 'high' and 'low' levels respectively for comparative purposes). 120 kg test blocks (250 mm diameter) were cast in resin-bonded sand at high (0.3%) and low (0.06%) phosphorus levels. Half the blocks were fast cooled from above 1000 °C by stripping before $1\frac{1}{2}$ hours. Chemical analyses of the test blocks are given in Table 1.

Mechanical properties were determined from the centre of each test block using conventional tensile tests and a bend ductility test[2]. The latter involves fracturing a 12.7 mm diameter bar in three point bending between 127 mm centres and determining the load-deflection curve. Typical load-deflection curves for ductile and brittle flake graphite irons are shown in Fig. 2. It can be seen that in the case of a brittle iron, once maximum load has been achieved the load falls more rapidly away to zero. It has been determined that the most suitable parameter for comparing the curve shapes is a/b (Fig. 2) this being the ratio of curve widths at half-peak height. High values of a/b therefore represent an embrittled iron.

Microstructures were determined from the ends of the fractured bend test specimens and typical structures are shown in Fig. 3. Ferrite contents are summarised in Table 2 together with the mechanical properties.

6. DISCUSSION

It can be seen from the microstructures (Table 2) that in the case of the low residual melts there is a substantial proportion of ferrite (85% in the slow cooled low phosphorus test block), the amount being reduced by the fast cooling practice. The high phosphorus test blocks were more pearlitic than the low phosphorus blocks but this is assumed to be due to the higher Mn level (0.73% as opposed to 0.46%). Of particular note was that the phosphide/carbide eutectic in the high phosphorus blocks was in the form of a fairly uniform dispersion with only a limited tendency to segregate to the cell boundaries.

The high residual melts were substantially pearlitic with only a trace of ferrite being evident in the slow cooled blocks. The phosphide/carbide eutectic in the high phosphorus blocks tended to be more segregated to the cell boundaries than observed in the low residual melts (Fig. 3).

Examination of Table 2 shows that in the case of the high residual melts, an increase in phosphorus content in the slow cooled test blocks produced a marked increase in brittleness, the a/b ratio increasing from 3.5 to 14.3. A similar increase was also found to occur in the fast cooled blocks. The fast cooled blocks were also embrittled compared with the slow cooled blocks, a/b ratios increasing from 3.5 to 7.1 and 14.3 to 17.5 for the low and high phosphorus blocks respectively. The most

brittle condition was represented by the high phosphorus fast cooled test blocks.

In the low residual melts the fast cooled high phosphorus test block again represented the most brittle condition, but it can be seen that the increase in cooling rate had a greater effect on properties than changes in phosphorus levels. In fact the matrix was more pearlitic in the case of the high phosphorus test blocks and this is known to increase brittleness, indicating that the real effect of phosphorus in this series of test blocks was only slight.

The effect of phosphorus on the high and low residual melts was substantially different. The most notable microstructural variation, other than ferrite content, was that the phosphide/carbide eutectic in the low residual melts was dispersed, whereas in the high residual melts it was more segregated to the cell boundaries. It is apparent therefore that high phosphorus levels will be most detrimental to cracking resistance when the phosphide/carbide eutectic is in the form of cell boundary networks which offer an easy path for cracks to propagate through the material. The present work has shown that this form is promoted by high trace element levels and fast cooling from 1000 °C.

The bend test results illustrate the importance of controlled cooling of moulds in the foundry. a/b ratios of low phosphorus casts were found to increase from 1.1 to 2.3 and from 3.5 to 7.1 with fast cooling for low and high residual melts respectively.

Comparison of foundry mould analyses (Section 4.1) with those of the experimental test blocks (Table 1) show that the moulds contained substantially higher levels of the pearlite and carbide stabilising elements, Cu, Cr and Mo than the 'high' residual test blocks. This would indicate that the moulds will be more prone to the deleterious effects of phosphorus and fast cooling than the test blocks. The cooling rate on solidification will also play an important role since the slower the solidification rate, as in ingot moulds compared with the test blocks, the greater the tendency for segregation at the head of the advancing solidification front and hence the greater tendency to form the deleterious cell boundary network of phosphide/carbide eutectic.

6. RELEVANCE TO MOULD PERFORMANCE

It is clearly apparent from Fig. 1 that the early failures of the WEU CB type moulds were due to phosphorus contents in excess of 0.12%. The mean mould performance dropped from 120 lives with low phosphorus (<0.07%) down to 68 lives for high phosphorus (>0.131%). The drop in life for the NEU 0T moulds, however, was only of the order of 10 lives for the same phosphorus levels.

It has been noted above that previous experience has claimed a beneficial effect of increased phosphorus levels. This clearly is not the case for these moulds. Since the experimental work showed the importance of stripping practice, temperature measurements at the foundry were made immediately after stripping the top plate which exposes the narrow end face of the ingot mould,

all ingot moulds being manufactured with the narrow end uppermost. It was found that the surface temperatures of the heavy base section of the WEU CB mould can lie between 890 and 980 °C. Surface temperatures on the exposed end face for the NEU OT moulds were below 750 °C. The much higher surface temperature of the WEU CB mould is due to its heavy base section (top as-cast) compared with the other mould types.

It is clear then that the necessary slow cooling conditions for high phosphorus moulds were not satisfied for the WEU CB mould for the then current cooling practice. Conversely, the cooling conditions for the NEU OT moulds more nearly approached the optimum and it is considered to be the reason for the relatively small decrease in performance for these mould types.

7. SUMMARY AND CONCLUSIONS

During the second half of 1974 a BSC ingot mould foundry operated with a burden which included up to 15% high phosphorus railway chairs. Mean phosphorus levels produced were approximately 0.12% but wide variations were observed between 0.035 and 0.264%. A slight deterioration in performance of up to 10 usages occurred for NEU OT moulds at high phosphorus contents, but severe problems were experienced with a WEU CB mould, the mean life falling from approximately 120 lives with low phosphorus (less than 0.07%) down to below 70 lives for phosphorus levels above 0.12%.

A series of test blocks was cast with 0.3 and 0.06% phosphorus with air cooling or sand cooling from 1000 °C to 700 °C in both high and low residual melts. In the high residual melts it was found that an increase in phosphorus levels markedly decreased the cracking resistance, this being attributed to the formation of a eutectic cell boundary network of phosphide/carbide eutectic. High phosphorus levels in the low residual melts were found to produce a lesser effect, the phosphide forming as a dispersion within the eutectic cells. It has been postulated that high trace element levels (carbide or pearlite stabilisers) promote the formation of the cell boundary network. It is considered that this effect will be enhanced in ingot mould production by the higher trace element levels than employed in the test blocks and also by the slower solidification rates.

Fast cooling of the test blocks from 1000 °C increased the brittleness of the test blocks and demonstrated the importance of slow cooling in the foundry. In the case of both high and low residual melts the most brittle condition encountered was a high phosphorus content coupled with a fast cooling rate from 1000 °C.

Previous experience has shown that phosphorus contents at this high level are not detrimental and may even be beneficial under slow cooling conditions in the foundry. Measurements carried out at the ingot mould foundry have shown that, in the case of the WEU CB mould, removal of the top plate can expose the heavy base section (top as-cast) at surface temperatures as high as 960 °C. The resulting rapidly cooled high phosphorus microstructure is therefore highly crack-sensitive, and resulted in the premature base cracking observed in this mould type.

The surface temperatures of NEU OT moulds are usually below 750 °C after removal of the top plates and this is considered to be the reason for the much decreased effect of phosphorus on the performance of this type of mould.

8. ACKNOWLEDGEMENT

The authors are grateful to Dr. K.J. Irvine, Manager, Sheffield Laboratories, for permission to publish this paper.

9. REFERENCES

1. Third Report of The Ingot Mould Sub-Committee, ISI Report No. 52, February 1955

2. Wilford, K.B. 'The Evaluation of the Cracking Resistance of Cast Iron by a Three Point Bend Test', British Foundryman, March 1980, pp.61-66

TABLE 1

CHEMICAL ANALYSIS OF TEST BLOCKS

Cast No.	Strip Temp. °C	C	Si	S	P	Mn	Mo	Cr	Sn	V	Ti	Ni	Al	Cu
RG202/A	>1000	3.75	1.80	0.012	0.058	0.46	0.001	<0.001	<0.001	0.010	<0.001	0.029	0.012	<0.03
RG202/B	< 700													
RG203/A	>1000	3.95	1.74	0.011	0.317	0.73	0.001	<0.001	<0.001	0.010	<0.001	0.024	0.018	<0.03
RG203/B	< 700													
RG742/A	>1000	3.61	1.30	0.031	0.058	0.89	<0.01	0.01	0.009	0.010	0.013	0.05	<0.005	0.06
RG742/B	>1000	3.55	1.18	0.027	0.325	0.79	<0.01	0.01	0.008	0.015	0.017	0.05	<0.005	0.07
RG743/A	< 700	3.60	1.31	0.025	0.059	0.91	<0.01	0.01	0.010	0.010	0.013	0.05	<0.005	0.07
RG743/B	< 700	3.63	1.23	0.024	0.310	0.94	<0.01	0.01	0.010	0.015	0.019	0.05	<0.005	0.07

TABLE 2

MECHANICAL PROPERTIES

Cast No.	Residual Level	Ferrite %	Strip Temp. °C	Phosphorus Level	UTS N/mm^2	Elong. %	Max. Load kN	a/b
RG202/B	Low	85	< 700	Low	39	0	1.0	1.1
RG202/A	Low	35	>1000	Low	59	1.4	1.2	2.3
RG203/B	Low	35	< 700	High	64	1.3	1.2	1.8
RG203/A	Low	5	>1000	High	85	0	1.6	2.9
RG743/A	High	< 5	< 700	Low	118	1.6	1.6	3.5
RG742/A	High	< 5	>1000	Low	144	0.6	1.9	7.1
RG743/B	High	< 5	< 700	High	177	1.3	2.1	14.3
RG742/B	High	< 5	>1000	High	213	OPM	2.5	17.5

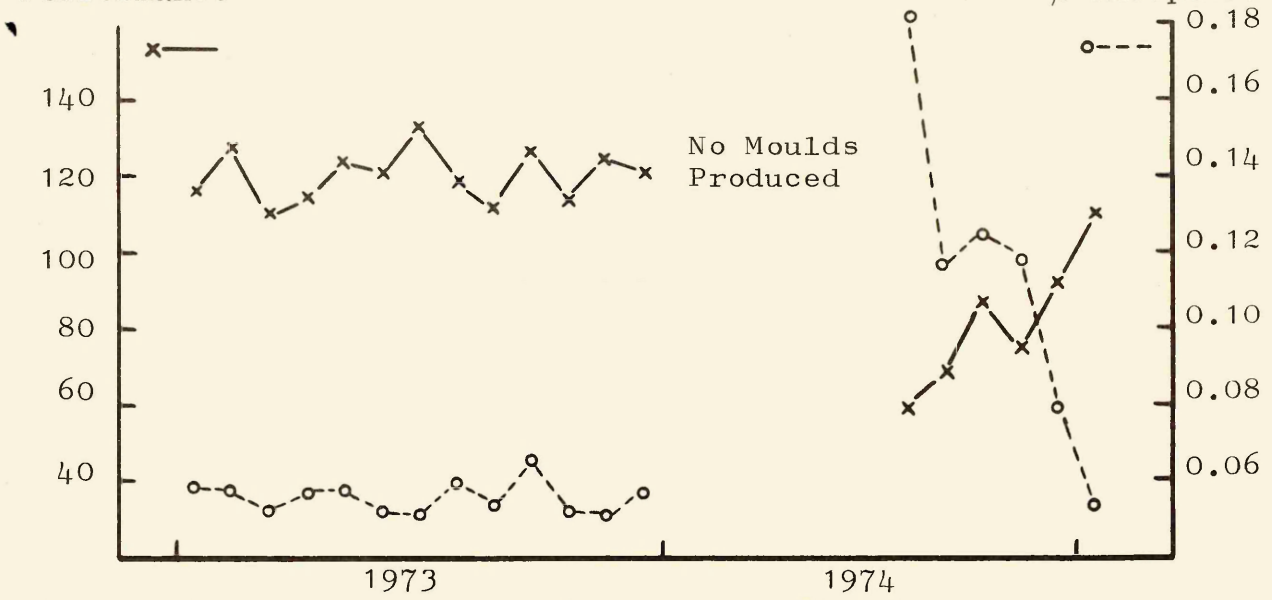


Fig. 1. Monthly Average Performance and Phosphorus Content of WEU CB Moulds at Steelworks S

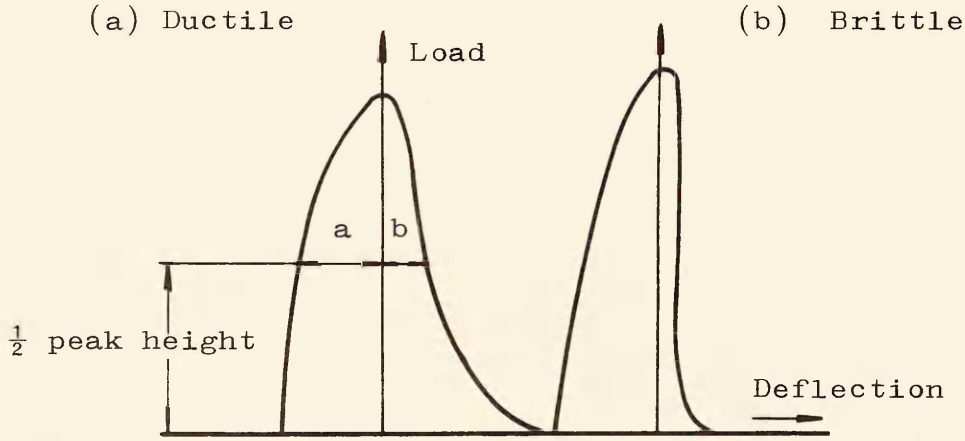


Fig. 2. Load-Deflection Curves for Ductile & Brittle Materials

(a) Low Residuals (b) High Residuals

Fig. 3. Effect of Residuals on Microstructure (x 100)

The evaluation of the cracking resistance of cast iron by a three point bend test

K B Wilford, B. Met, (Associate Member)

Three point bend tests have been carried out on experimental and commercially produced cast iron using 12·5 mm diameter specimens at 127 mm centres. It has been found that the load-deflection curve may be analysed in terms of the areas before and after maximum load to determine the cracking resistance of the iron. These areas can be related to the energies to initiate and to propagate fracture respectively. It has been shown that brittle irons show a low propagation energy although the initiation energy may increase because of an increase in strength, so that the ratio of the two energies provides a useful comparison of brittleness. It has been shown that brittleness increases as the flake graphite size is reduced, or the pearlite content is increased. The presence of carbides in modified graphite iron is also detrimental in this respect. The test offers a simple and cheap alternative to conventional fracture toughness testing.

1. INTRODUCTION

The design criterion of material toughness is becoming increasingly more important in the metallurgical industry but is not widely used for the specification of iron castings. However, the cracking resistance of cast iron can be important and this is particularly true for ingot moulds. The performance of an ingot mould in service is determined mainly by its cracking resistance, since almost all premature mould failures occur by cracking. Although mould cracking is affected by mould design and steelworks usage conditions, metallurgical factors are known to play an important part.

Grey cast irons fracture in a tough or brittle manner depending on the analysis and microstructure. However, the quantification of the ductility of cast iron is difficult by conventional mechanical tests since in the tensile test the elongation or reduction in area is low—of the order of 1 or 2% even for a relatively ductile flake graphite iron. Similarly, the Charpy impact values are of the order of 5J or below on unnotched specimens.

It has been suggested[1] that the toughness of cast iron can be characterised by a modified fracture toughness test using an unfatigued CKS type specimen. However, these test pieces are expensive to machine, the tests require non-standard testing equipment and the results require skilled interpretation.

A simpler and cheaper test can be carried out using a three or four point bend test and analysing the data in terms of the areas before and after maximum load, on a load-deflection plot. These areas may be loosely equated with the energy to initiate fracture and the energy to propagate fracture through the specimen, although it has been shown using the acoustic emission technique that crack propagation can occur considerably before maximum load is attained[2].

The technique of comparison of areas before and after maximum load has been used in the testing of iron[3] and also in steels to determine the susceptibility to lamella rtearing using a test piece similar to a modern CKS specimen.[4] This test has become known as the US Navy Tear Test and has also been used to assess austenitic spheroidal graphite irons[5]. Three point bend tests have also been used more recently[6] in assessing the susceptibility of welded steel to lamellar tearing.

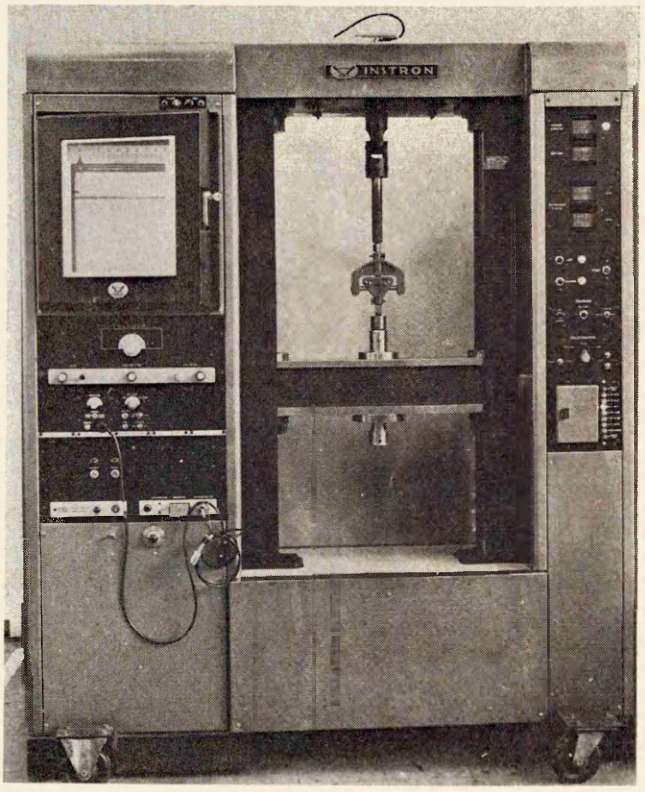


Fig. 1—Instron testing machine with bend test attatchment

Work is currently in hand at the author's laboratory to assess the effect of production variables on ingot mould life. Bend tests are being carried out on various test blocks and the results have been analysed in order to assess the suitability of the bend test as a means of determining cast iron cracking susceptibility. The paper describes the basic test

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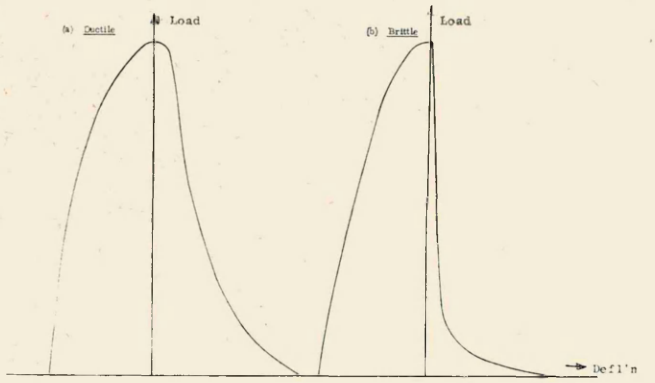


Fig. 2— Load deflection curves for ductiles and brittle materials

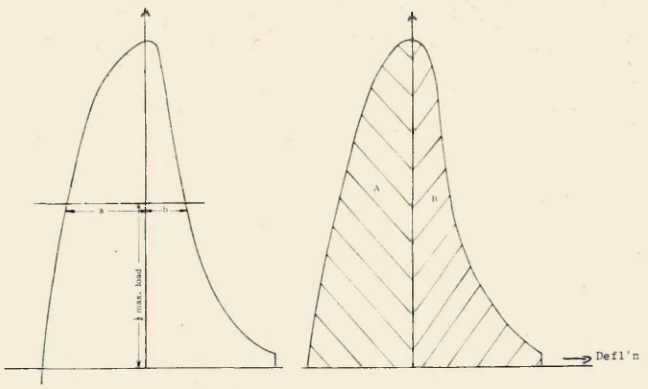


Fig. 3—Determination of bend test parameters

and shows how the test can be used to rank irons in terms of their ductility or brittleness.

2. TESTING METHOD

A cast iron bend test attachment [supplied by Industrial and Educational services] was used with a cylindrical bar specimen of 150 mm length × 12.5 mm diameter, being loaded under three point bending on 127 mm centres. The attachment was connected to an *Instron* universal testing machine to produce a continuous load-deflection curve. A test in progress is shown in Fig. 1. The *Instron* crosshead movement was set at 10 mm/min with a chart speed of 200 mm/min. Three bend test samples were tested for each treatment/analysis condition and the results averaged. The testing equipment is also suitable for use on miniature test pieces.

3. ANALYSIS OF LOAD-DEFLECTION CURVES

Typical load-deflection curves for ductile and brittle flake irons are shown in Fig. 2a and b respectively. It can be seen that in the case of a brittle iron, once maximum load has been attained unstable crack propagation occurs, through the specimen. This occurs since the energy required to propagate is low so that only a small amount of energy is required in addition to that stored in the specimen as elastic strain energy.

The parameters measured to describe the shape of the load-deflection curves are given in Fig. 3. These are as follows:—

1. Maximum load.
2. Ratio of the width of the curve before and after maximum load at half maximum load, a/b.
3. Energy to initiate a crack, area A.
4. Energy to propagate the crack, area B.
5. Ratio of initiation and propagation energies, A/B.

It can be seen, with reference to Fig. 2. that high values of a/b and A/B are related to brittle material.

4. EXPERIMENTAL MATERIAL AND RESULTS

4.1 Effect of graphite size

Bend test samples were prepared from a series of test blocks of similar melt analyses but with differing titanium treatments to produce a range of graphite sizes. Microstructures of the five samples are given in Fig. 4. Load deflection curves for the specimens are shown in Fig. 5. and the measured bend test parameters in Table I.

It can be seen that as the flake size is reduced from coarse to very fine undercooled graphite, there is an increase in maximum load but this is accompanied by a large decrease in propagation energy. The parameters a/b and A/B both show a large increase with decreasing flake size indicating a marked increase in brittleness.

The embrittling effect of undercooled graphite has long been known, for instance in the case of 6% silicon irons although this is complicated by the solid solution embrittling due to the high silicon content. An increase in graphite size may be brought about in the foundry by increases in carbon, silicon and titanium levels and this is known to increase the cracking resistance of ingot mould iron.

4.2 Effect of matrix microstructure

Bend test samples were prepared from a fully pearlitic iron test block. Offcuts were then given sub-critical anneals at 700°C for 48 and 96 hours to produce 60 and 100% ferrite respectively and further bend test samples prepared. Resultant load-deflection curves are given in Fig. 6 and the derived bend test parameters in Table II.

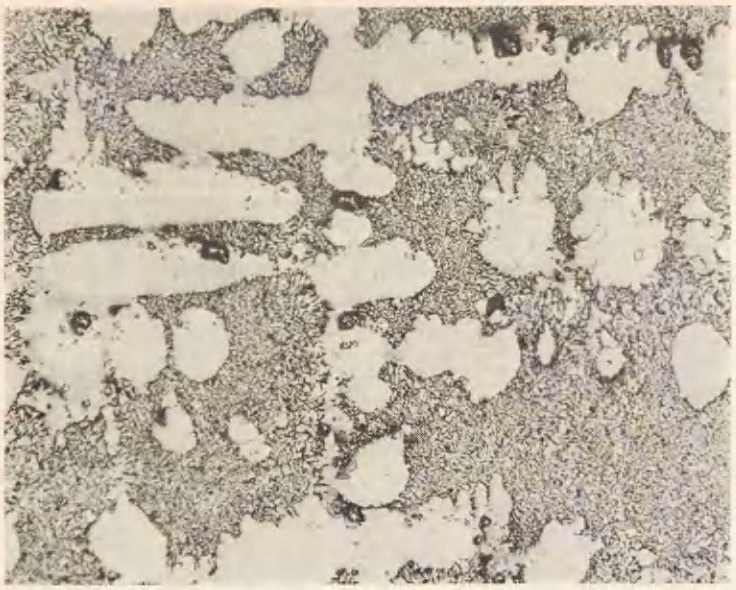
As the amount of ferrite is increased, the maximum load decreases and this is accompanied by an increase in propagation energy (area B). The a/b and A/B ratios both decrease indicating an increase in cracking resistance.

Table I Effect of graphite size on bend test parameter

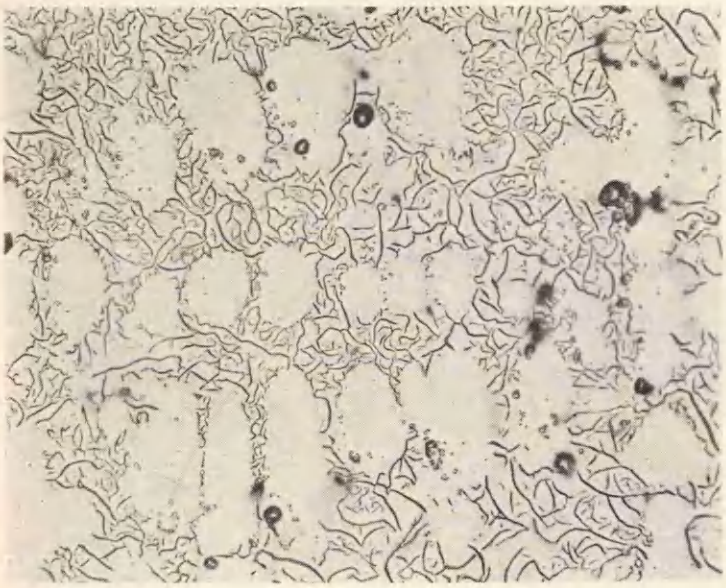
Sample No.	*Max. Load, kN*	*a/b*	*A*	*B*	*A/B*
1 (Fig. 4)	3·5	~35	2·2	0·01	~200
2 ,,	2·1	~20	1·3	0·2	6·5
3 ,,	1·5	9·2	1·2	0·5	2·4
4 ,,	1·4	2·1	1·1	0·8	1·4
5 ,,	1·4	2·3	1·0	1·1	0·9

Table II Effect of percentage pearlite on bend test parameters

Pearlite, %	*Max. Load, kN*	*a/b*	*A*	*B*	*A/B*
100	1·6	2·9	1·0	0·7	1·4
40	1·3	2·7	1·0	0·8	1·3
0	1·3	1·7	1·0	1·0	1·0



Sample 1



Sample 2

Sample 3

Sample 4

Sample 5

Fig. 4—Graphite morphologies ×100

This again correlates well with common experience in that a reduction in the pearlite content results in increased fracture resistance. The effect however, appears to be smaller than for variations in graphite size.

4.3 Effect of graphite morphology

Bend test samples were prepared from ferritic quasi-flake and spheroidal graphite test blocks and from an ingot mould containing compacted graphite (typical of high nitrogen levels in heavy sections) with approximately 50% ferrite.

Typical load-deflection curves are shown in Fig. 7 with a ferritic flake iron for comparison and the bend test parameters are given in Table III. The SG iron sample was so ductile that fracture was not initiated before the test had to be discontinued. The load-deflection curve was similar in shape to that observed in the quasi-flake iron up to the point when the test had to be stopped.

The a/b and A/B ratios are numerically greater than those for flake iron which, based on observations above, would lead to the conclusion that modified graphite irons are more brittle than flake irons. This, clearly, is untrue and is seen to arise as a consequence of the much greater initiation energies for the modified irons.

The bend test does not appear, therefore, to be useful for comparing irons of differing graphite morphologies. However, the ductility within groups can still be compared. This is illustrated in Figs. 8 and 9 which are load-deflection curves from ferritic quasi-flake moulds with and without carbide and from compacted graphite moulds having failed normally and prematurely due to major cracking, respectively. It can be seen in both cases that the more brittle sample exhibits both reduced initiation and propagation energies with an overall increase in a/b and A/B ratios.

5. APPLICATION OF THE BEND TEST TO INGOT MOULDS

5.1 Premature cracking in large slab moulds

Bend test samples were taken from three prematurely failed moulds and from a high life mould and compared. The results are given in Table IV and the bend test curves shown in Fig. 10. It can be seen that the high life mould possesses much lower a/b and A/B ratios indicating a higher resistance to cracking than the prematurely failed moulds. Adjustments were made to the chemical analysis of subsequent moulds in order to improve cracking resistance and the incidence of premature failures has been reduced.

Table III Effect of graphite morphology on bend test parameters

Condition	*Max. Load, kN*	*a/b*	*A*	*B*	*A/B*
Flake	1·3	1·7	1·0	1·0	1·0
Compacted	2·6	2·6	3·6	2·0	1·8
Quasi-flake	3·5	3·5	7·5	2·9	2·6

Table IV The effect of bend test parameters on ingot mould life

Mould life	*Max. Load kN*	*a/b*	*A*	*B*	*A/B*
120	1·0	0·5	0·6	1·8	0·3
24	1·2	0·9	0·4	0·8	0·5
15	1·6	2·5	0·8	0·5	1·6
2	1·3	1·7	1·0	1·1	0·9

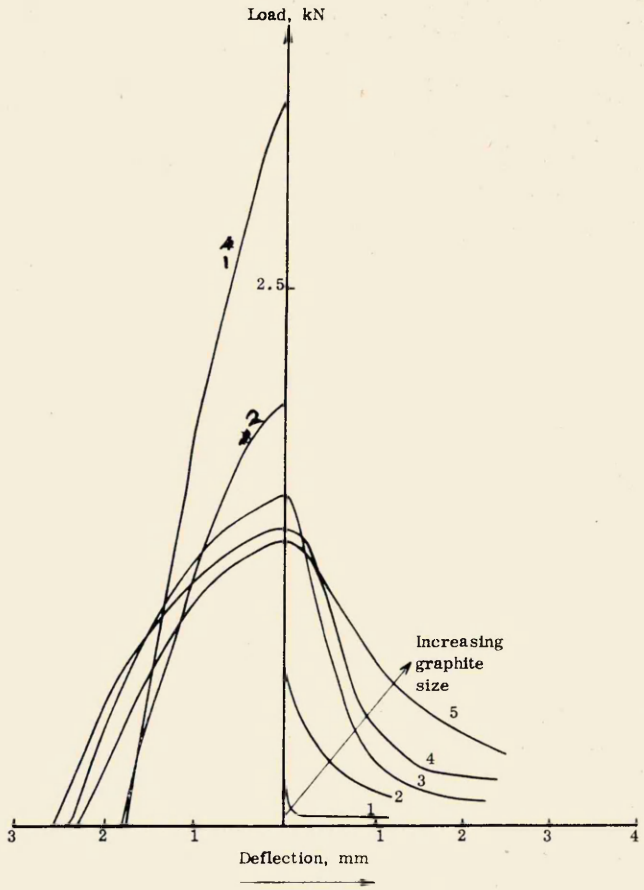


Fig. 5—Effect of graphite size on load-deflection curves

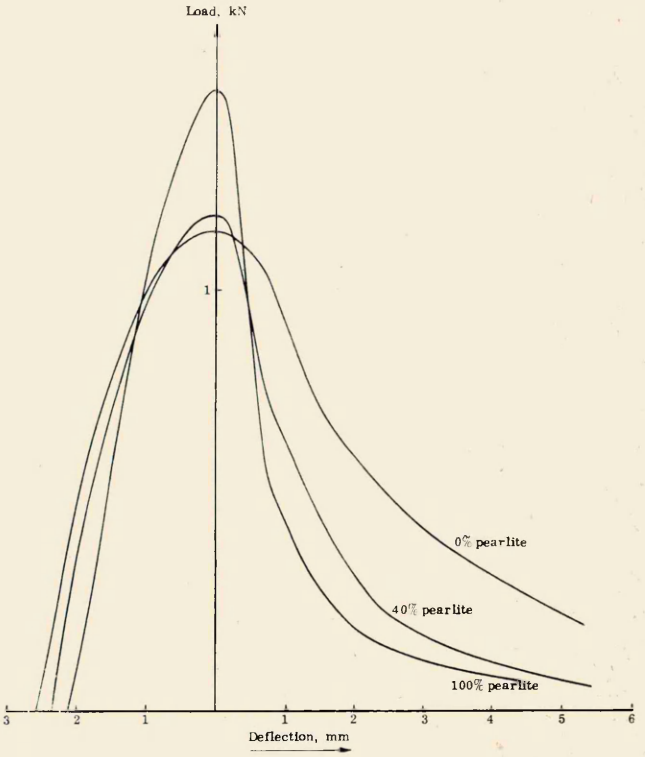


Fig. 6—Load-deflection curves showing effect of amount of pearlite

5.2 Moulds failing due to Base Cracking

Moulds were failing prematurely by cracking through the heavy base section (top as cast). Bend test samples were taken from the base and also 150 mm up the wall. Typical load deflection curves are shown in Fig. 11. It can be seen that there is localised embrittlement of the base. A foundry investigation revealed that this was caused by removal of the top plate whilst the base was above 950°C. The stripping practice was improved and the incidence of base cracking was reduced.

6. SUMMARY AND CONCLUSIONS

Three-point bend tests of cast iron have been carried out using a 12·5 mm diameter bar at 127 mm centres. Two parameters have been found to be capable of adequately describing the cracking susceptibility of the material.

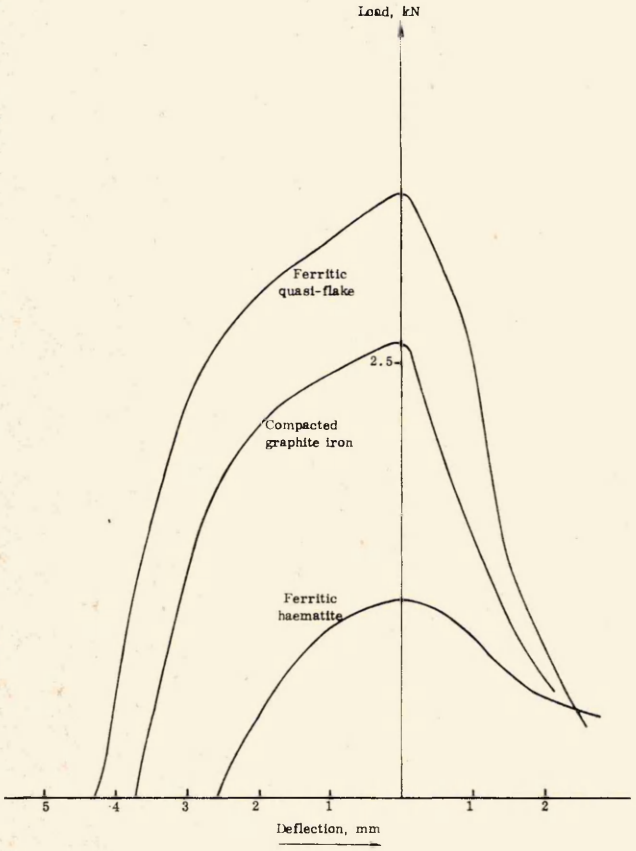


Fig. 7—Load-deflectian curves for ferritic haematitic and quasi-flake irons

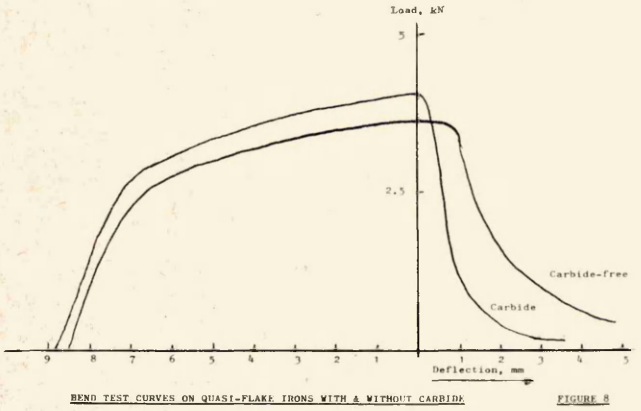


Fig. 8—Bend test curves on quasi-flake irons with and without carbide

(a) Ratio of widths at half peak height, a/b
(b) Ratio of areas before and after maximum load, A/B

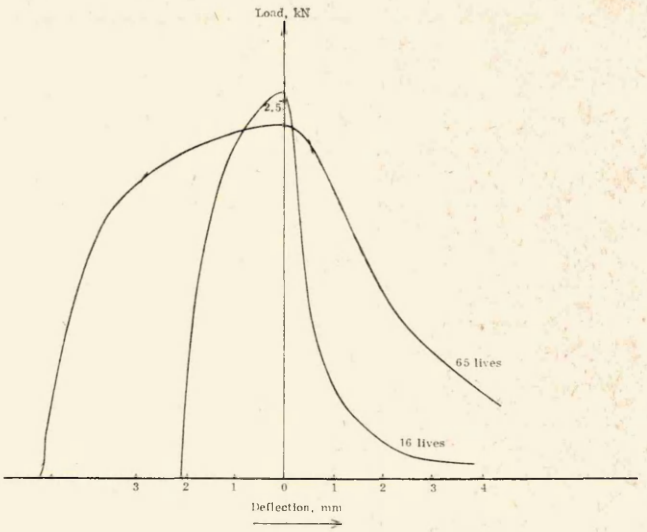


Fig. 9—Bend test curves on compacted graphite moulds

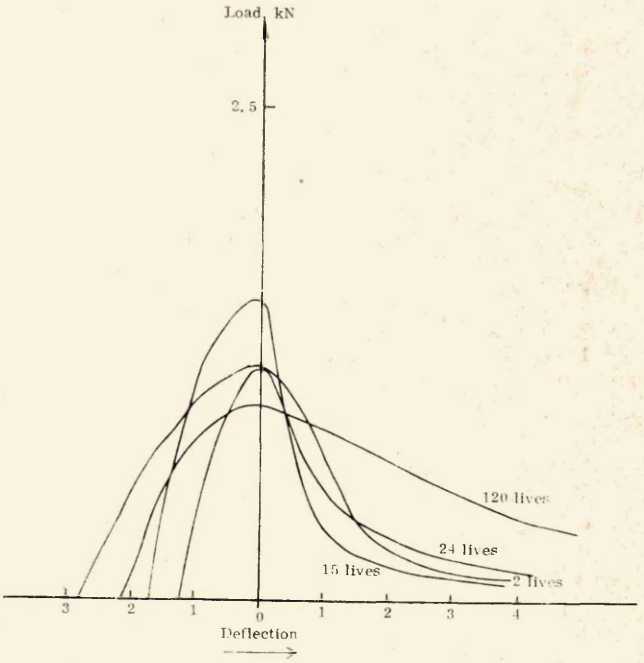


Fig. 10—Bend test curves of flake graphite moulds

The area up to maximum load may be equated with the energy to initiate fracture and the area after maximum load with the energy to propagate the crack through the ligament. Both ratios show an increase as the material becomes more brittle. This has been obtained by refining the graphite size, increasing the amount of pearlite and the presence of carbide in quasi-flake iron. Of the ratios, the ratio of widths at half peak height (a/b) is the easiest and quickest to measure and may be suitable for routine testing.

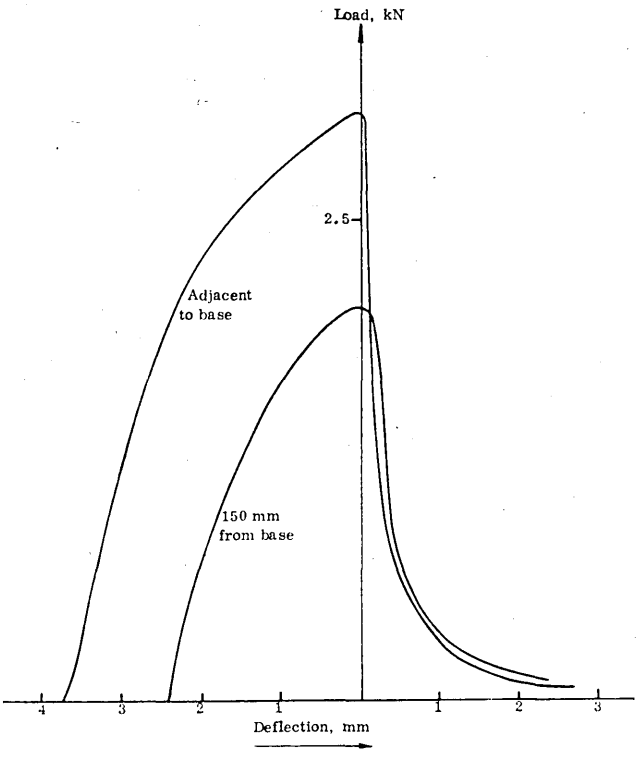


Fig. 11—Load-deflection curves from mould failed by base cracking

Modification of graphite morphology, for example by magnesium treatment, results in a modification of the load-deflection curve such that initiation energies and maximum loads are increased. Comparison of these curves with those from flake irons using the ratios described above is not possible. It has been demonstrated, however, that the cracking resistance of irons of the same graphite morphology may be compared using the bend test. Examples of the use in diagnosing ingot mould failures have been given.

The bend test appears to offer a simple and cheap test both for quality control purposes and as a research tool for predicting cracking tendency.

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