

THE CAST STRUCTURE OF HIGH-SPEED STEEL

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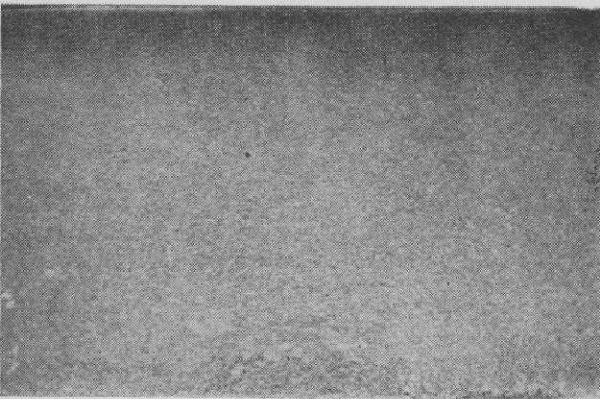
Abstract

The cause and effects of the formation of coarse cellular carbides in high-speed steels are reviewed and attention is drawn to possible methods of investigating the problem and perhaps eliminating it.

Introduction

ONE of the greatest problems encountered in the manufacture and use of high-speed steel is the formation during casting of a gross cellular carbide network, particularly near the centre of large ingots. A section from an ingot with such defects is shown in Fig. 1, in which it is seen that there are quite large areas of carbide. Microscopic examination of this ingot showed that near the edge the structure was quite fine-grained (FIG. 2) whereas at the centre the grain size is large (FIG. 3) and there are large colonies of eutectic.

Edge of Ingot



Centre of Ingot

FIG. 1 — HALF TRANSVERSE SECTION OF HIGH-SPEED STEEL INGOT SHOWING MASSIVE CARBIDE AT CENTRE. $\times \frac{1}{2}$

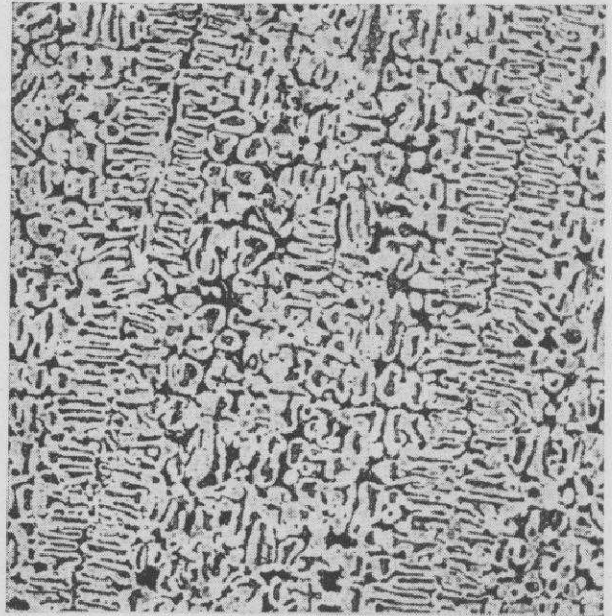


FIG. 2 — STRUCTURE NEAR EDGE OF INGOT SHOWN IN FIG. 1. $\times 100$

Also there is a marked coring effect in the primary dendrites themselves. It is the eutectic colonies shown in Fig. 3 which give rise to the segregated appearance of Fig. 1. The phenomenon of poor carbide distribution has often been referred to as carbide segregation: this is, in fact, rather a misnomer, since it is confirmed by numerous methods^{1,2} that the carbon content of high-speed steel ingots is practically uniform throughout, and also that there is no appreciable difference in the proportion of carbide (eutectic) in the structure at the edge and centre.

The elimination of the gross eutectic network is essential if a good quality high-speed steel tool is to be produced. The carbide in high-speed steel is, however, very complex and, unlike cementite in ordinary engineering

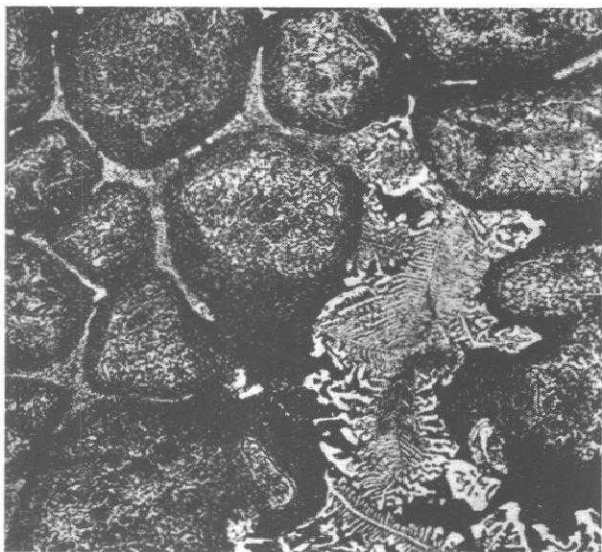


FIG. 3 — STRUCTURE NEAR CENTRE OF INGOT SHOWN IN FIG. 1. $\times 100$

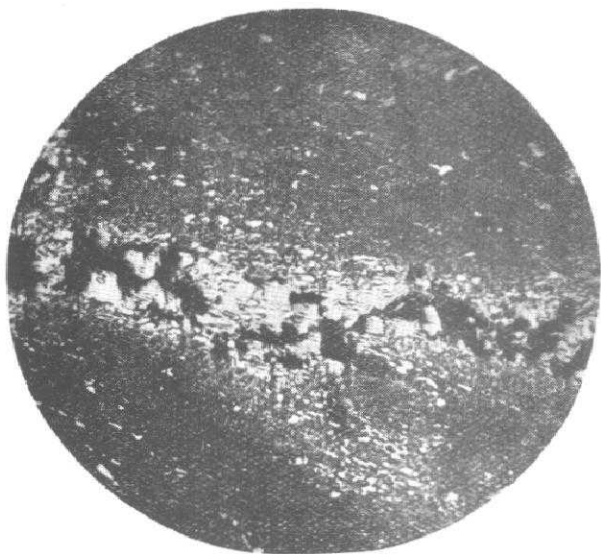


FIG. 4 — BURST IN HIGH-SPEED STEEL

steels, is extremely refractory and comparatively insoluble under ordinary annealing conditions. The only known effective way of breaking down the carbide network is by extensive hot-working involving some 90-95 per cent reduction of area. It is here that the first troubles are encountered, and in ingots with poor carbide distribution extreme care is required if centre bursting (see FIG. 4) is to be avoided.

The alteration in structure by hot-working is shown in Figs. 5 and 6. Owing to the very low solubility of carbide (complex eta iron-tungsten carbide predominates in high-speed steel) even at very high temperatures, the break-up of the eutectic carbide network is brought about mainly by mechanical means. Even in the finished bar, the effects of good and bad carbide distribution are still discernible. Fig. 7³ shows the comparison of two high-speed steel bars one of which is reasonably homogeneous and the other contains marked carbide stringers. The unnotched izod impact values of these steels, after hardening and double tempering, are given in Fig. 8 and show that the homo-



FIG. 5 — HIGH-SPEED STEEL INGOT AFTER FORGING. $\times 100$

geneous steel has a much higher toughness than the banded steel.

There are limits to the effectiveness of hot-working. The gross carbide network becomes progressively worse as the size of the

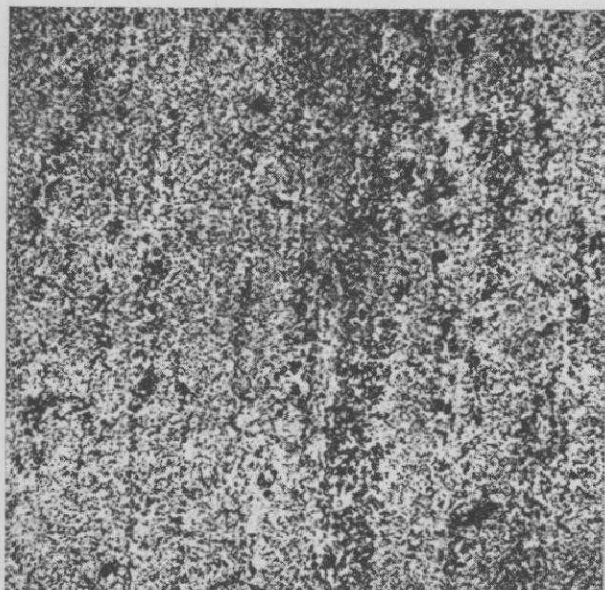


FIG. 6 — AS FIG. 5, WITH GREATER DEGREE OF WORKING. $\times 100$

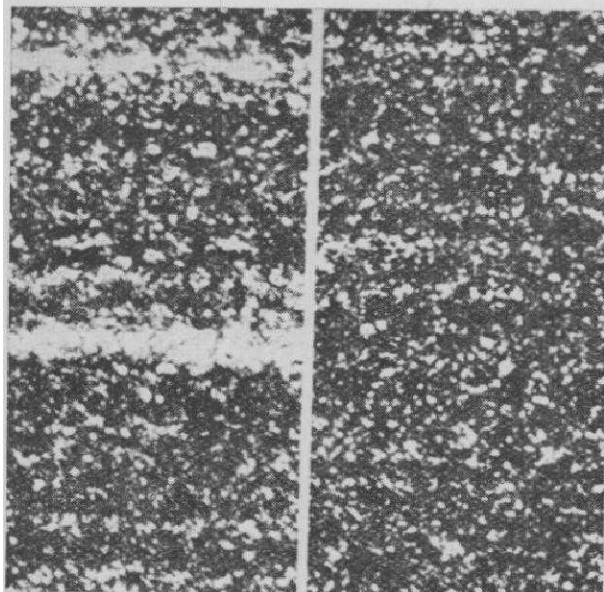
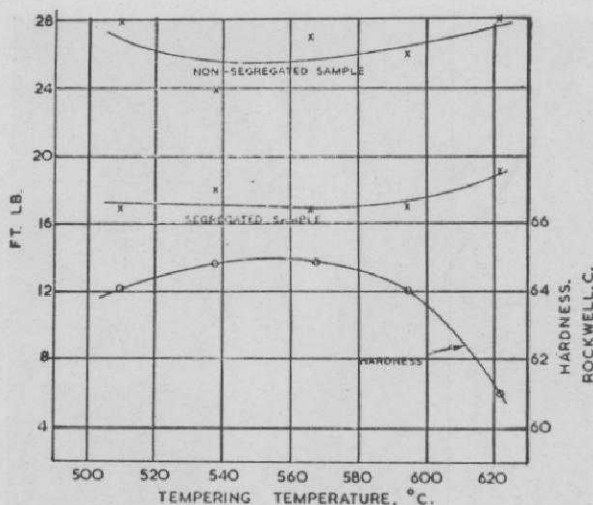


FIG. 7 — COMPARISON OF HIGH-SPEED STEEL BARS SHOWING GOOD (ON THE RIGHT) AND BAD CARBIDE DISTRIBUTION



COMPARISON OF UN-NOTCHED IZOD TESTS ON SEGREGATED, AND NON-SEGREGATED HIGH SPEED STEEL.

FIG. 8 — PROPERTIES OF BARS SHOWN IN FIG. 7

say, a 6 in. diameter cutter blank, it can be seen that to achieve a 90-95 per cent reduction of area in hot-working it would be necessary to start with a 24 in. ingot. This is practically the limit in size of commercial high-speed steel ingots, and it will be appreciated that new requirements by engineers for 9 in. and even 12 in. diameter blanks are inevitably leading to greater difficulty in producing sound blanks free from the so-called centre weakness. In the case of such tools as milling cutters it is, of course, practicable to make large blanks by upsetting, but this is not possible for blanks for many tools such as hobs. Fig. 9 shows a typical drill which has failed due to poor carbide distribution (probably with insufficient reduction).

Almost from the inception of high-speed steels some 50 years ago, steel-makers have endeavoured to overcome the difficulty by attempting to refine the cast structure in the ingot. It has proved to be a most intractable problem and little progress has been made apart from relatively minor improvements brought about by applying stricter control to steel-making processes. There is good evidence, however, that the only

ingot increases, and in ingots greater than about 14 in. sq. it is so pronounced that no convenient amount of hot-working will entirely eliminate the cast structure. If one considers a quite normal requirement for,

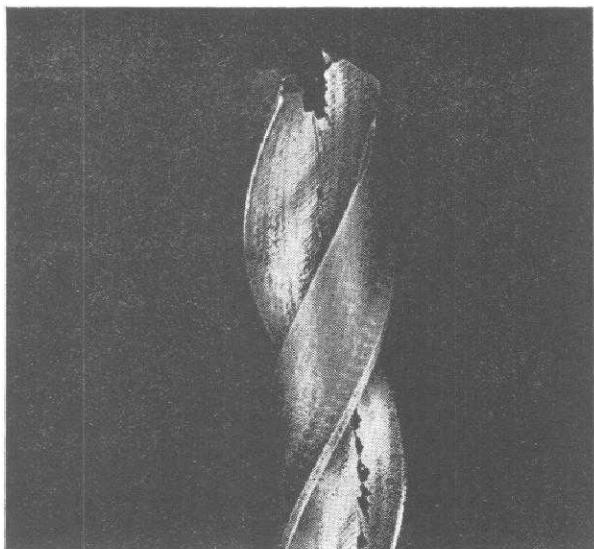


FIG. 9 — FAILURE IN HIGH-SPEED STEEL DRILL DUE TO POOR CARBIDE DISTRIBUTION

completely satisfactory answer to the problem would be the discovery of some method of modifying the cast structure of high-speed steel. B.I.S.R.A. have, therefore, undertaken to examine the question again in the light of more recent theories of nucleation and growth of crystals and are employing new or improved techniques which have become available.

The Constitution of High-speed Steel

High-speed steels are complicated alloys of Fe, W, Cr and V, containing usually between 0.5 and 0.8 per cent carbon. The basic alloy composition which has found universal favour is still the one introduced by Taylor and White in 1900, namely W, 18; Cr, 4 and V, 1 per cent (18-4-1); but there are many variations, the most important of which are those in which molybdenum replaces or supplements tungsten. In a general consideration of the constitution and behaviour of high-speed steels the many variations in composition may be largely ignored and attention concentrated

on the basic 18-4-1 type. It has been customary to consider high-speed steels as approximating to ternary iron-tungsten-carbon alloys by supposing that chromium, vanadium and molybdenum behave similarly to tungsten. The tungsten equivalence of each of these elements differs from unity and it has been claimed⁴ that 1 per cent of Cr, Mo, or V is equivalent respectively to about 0.5, 1.5 or 5.0 per cent of tungsten. Goldschmidt⁵ does not entirely agree with this, particularly in regard to chromium, which does not appear to behave in quite the same manner as tungsten. However, as a first approximation an 18-4-1 high-speed steel can be considered as being equivalent to a 25-26 per cent tungsten-iron-carbon alloy.

Much work has been done on the constitution of ternary Fe-W-C alloys, but very little of this is complete or confirmed. The most widely accepted data were provided by Takeda^{6,7} in 1930 and 1931. On the basis of these data a section through the orthodox ternary Fe-W-C system at 25 per cent W and parallel to the Fe-C side of the diagram has been constructed. Another variation of the diagram was published by Grossmann and Bain⁴, who prepared a pseudo-binary diagram of iron against complex carbide. Later Murakami and Hatta⁸ made an attempt to give a closer approximation to the true state of affairs by preparing a binary section through a quaternary Fe-Cr-W-C system at 18 per cent W and 4 per cent Cr. Subsequently Kuo⁹ has proposed a modification of Murakami and Hatta's diagram. The version according to Kuo is reproduced in Fig. 10. No account is taken of vanadium in this diagram, but Kuo suggests that, from a consideration of the known behaviour of this element, adequate correction for its presence is obtained by moving the respective points on the diagram slightly towards higher carbon contents.

In the fully worked and annealed condition, an 18-4-1 alloy containing 0.7 per cent C

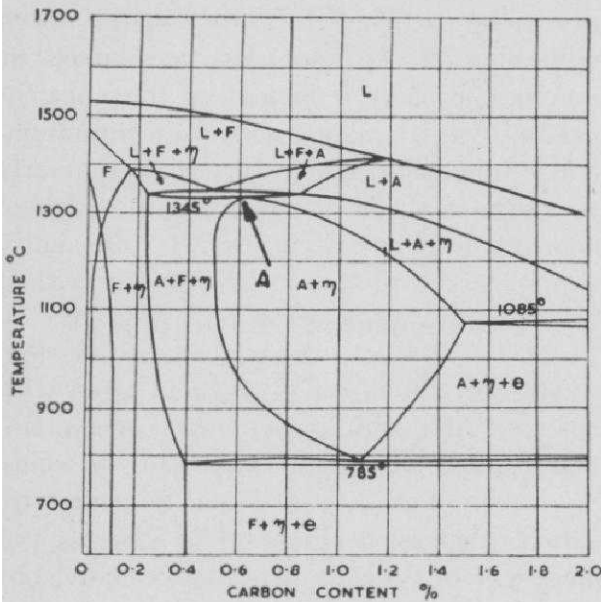


FIG. 10 — CONSTITUTION DIAGRAM OF THE Fe-W-Cr-C SYSTEM. SECTION AT 18 PER CENT W AND 4 PER CENT Cr (AFTER MURAKAMI AND HATTA, MODIFIED BY KUO). A — AUSTENITE; F — FERRITE; L — LIQUID; η — $(Fe.W)_6C$; θ — Fe_3C

consists essentially of ferrite and eta carbide. The latter is predominantly a complex carbide having the formula Fe_4W_2C or Fe_3W_3C — perhaps better expressed as M_6C or $(Fe.W)_6C$. It will be seen from Fig. 10 that theoretically the phase changes occurring during solidification of such an alloy will, allowing for the presence of vanadium, be indicated by a vertical line through A, or a line slightly to the left or right of A depending on the exact composition.

The reactions at equilibrium vary slightly according to the exact position on the diagram, but the conditions in practice are such that equilibrium is never attained and the following phases are present in the solidified steel:

- (1) Areas which were of delta iron at high temperature and which appear as islands of delta euctoid — i.e. finely dispersed carbide.
- (2) Bands surrounding the islands — believed to be austenite or martensite formed by transformation.

- (3) Colonies of eutectic occupying the interstices between the crystals and constituting the cellular network.
- (4) Isolated carbide particles — possibly vanadium carbide.

Freezing of Ingots

In the molten condition, high-speed steel is generally believed to be a practically uniform liquid throughout. The various elements present in addition to iron, such as C, Mn, Si, S, P, Cr, W, Mo and V, are essentially all dissolved. From time to time the view has been expressed that this is not entirely true and there have been suggestions that under certain conditions liquid high-speed steel will separate into two immiscible layers of different alloy content. No concrete evidence has ever been produced to support these views. However, whether or not the steel is completely uniform in the liquid state it certainly loses uniformity upon freezing as shown in Fig. 3, from which it is clear that marked coring or microsegregation exists within the primary dendritic regions quite apart from the major differences in composition between the primary and eutectic phases.

The cast structure may be relatively coarse or fine, depending on the rate of solidification in the mould. This in turn depends on the pouring temperature and the rate of cooling in the mould. The use of liquid steel pyrometers enables pouring temperatures to be controlled and reproduced to a high degree of accuracy, but it is not so easy to alter the rate of cooling in the mould. With smaller ingots some slight acceleration of cooling may be achieved by the use of copper moulds, or by water-cooling the mould: the chilling effect is, however, offset to some extent due to the fact that the initial chilled layer is immediately separated from the mould wall due to shrinking and leaves an air gap.

In any event the effect of chilling is negligible at the centre of large ingots —

which is the position where the refining effect is most needed if any beneficial result is to be obtained. We must, therefore, consider other means of inducing a fine structure and it is helpful in this connection to consider the conditions governing the formation of crystals from the melt.

Refinement of the Structure

If we consider the free energy changes as a crystallite forms, we find that the total free energy change ($\Delta G'$) is the sum of two terms — first, a positive term (*see* FIG. 11)¹⁰ $\Delta G'_2$ representing the energy of formation of the new interface and roughly proportional to the square of the diameter of the crystallite, and, secondly, a negative term $\Delta G'_1$ representing the energy of the phase change at a given degree of undercooling, and proportional to the cube of the crystallite diameter. The resulting total free energy change $\Delta G'$ thus shows a maximum corresponding to a certain critical particle size. Now, if a solid particle smaller than the critical size is formed in the melt, it will be unstable since any increase in its size would be accompanied by an increase in free energy and will tend to redissolve in the liquid. However, if a particle does form and grows

to a size in excess of the critical value, it will be stable and can grow with a decrease in free energy. Any reduction in the 'energy barrier' will, therefore, lead to an increase in the frequency with which stable crystallites form in the melt, and thus to a refinement of the grain size in the solid metal.

Homogeneous Nucleation

The dotted curves in Fig. 11 show the effect on $\Delta G'_1$ and $\Delta G'$ of undercooling the melt: $\Delta G'_1$ decreases to $\Delta G''_1$, thus reducing the value of $\Delta G'$ to $\Delta G''$ and reducing the critical nucleus size for stable growth and thereby increasing the probability of crystallite formation. $\Delta G'_2$ remains approximately constant during undercooling.

This indicates that refinement of structure by undercooling is possible when the nucleation is homogeneous — that is when each nucleus is formed from the melt and there is no effect due to foreign substances. In practice it is very seldom that nucleation is homogeneous¹¹, and there are inevitably effects due to mould walls, etc., which prevent undercooling and induce some measure of heterogeneous nucleation. Nevertheless, the principle of undercooling can be applied to some extent by rapid cooling, and this can best be achieved if a casting temperature as low as possible is used — as mentioned previously, however, this effect is negligible at the centre of a large ingot.

Effect of Heterogeneous Nucleation

An alternative method of reducing the energy barrier is by the addition of foreign particles to the melt. Under favourable conditions, the presence of foreign nuclei can lead to a reduction in $\Delta G'_2$ (*see* FIG. 12), thus once again causing a reduction of $\Delta G'$ to $\Delta G'''$ with a similar effect in increasing the frequency of crystal formation.

Attention is being focussed on the method of nucleation as being the most likely process

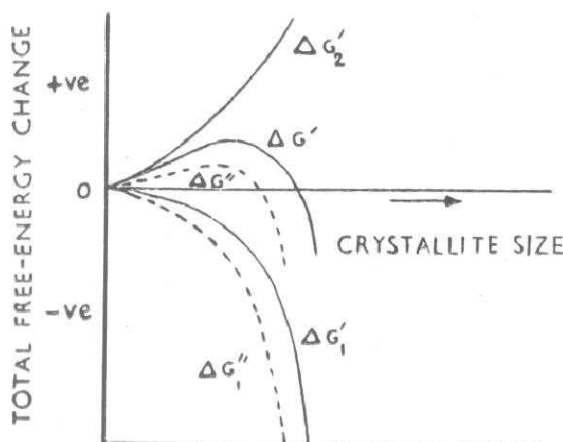


FIG. 11 — FREE ENERGY CHANGES DURING SOLIDIFICATION

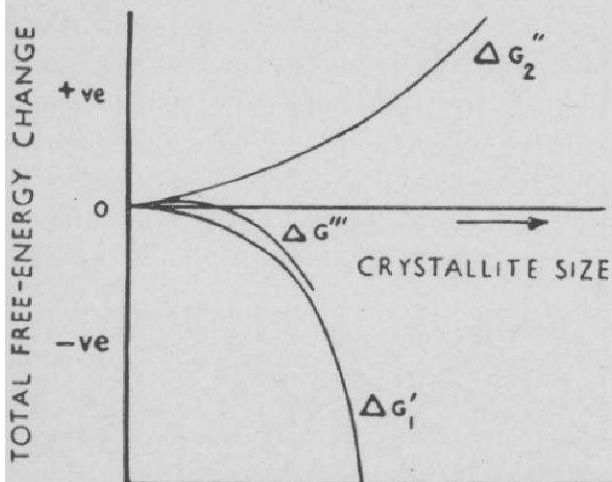


FIG. 12 — FREE ENERGY CHANGES DURING SOLIDIFICATION

to produce grain refinement in the present instance. In other fields some remarkable successes have been achieved by the application of this method¹²⁻¹⁴ and much work has been done in connection with the theory of the process, and to determine the conditions necessary for a particle to act as a nucleus.

The first and most obvious condition is that the substance selected as an inoculant must be capable of existing in the melt in the form of small solid particles at the solidification temperature. It is also essential that the dispersion of particles is satisfactory, and it is an advantage in this connection if the inoculant can be formed in the melt by reaction between a component already present and an added substance. Recent work by Thury¹⁵ has been described in which nuclei for refinement of aluminium alloys are formed by additions such as potassium boron fluoride, which reacts in the melt to produce aluminium boride: the reaction is exothermic and does not lower the temperature of the melt.

The second condition is that the nucleus must be wetted by the liquid metal, and a third is that attractive forces should exist between atoms of the solid metal and the nucleus; this factor must not, however, be

too great, or solution of the nucleus would occur by the formation of intermetallic compound.

The remaining condition is that the molten metal should be able to solidify on the artificial nucleus, or nucleation catalyst, just as if the nucleus were in fact a particle of the solid metal. In nature numerous examples are found where a crystal has grown on a crystal of another substance with a specific orientation relationship. This phenomenon is termed oriented overgrowth or epitaxy, and the present subject — in which crystallites are supposed to form and grow on an artificial nucleus in a similar way — has been referred to as isotaxy, or epitaxial nucleation. The conditions under which such a process can take place have been the subject of much attention and several mathematical treatments have been published¹⁶.

The criterion which is considered to be of major importance is that for nucleation to take place, the crystal structure of the nucleus and of the solidifying metal should be such that there is little or no strain at the interface between the two substances in the compound crystal formed. The concept of lattice registry, or misfit, is used as a measure of the amount of strain at the interface, and is defined as¹⁶:

$$\delta = \frac{\Delta a}{a_0}$$

where a_0 = lattice spacing in metal crystal

$a_0 + \Delta a$ = lattice spacing in metal nucleus.

This may also be expressed as a percentage, where

$$\% \text{ misfit} = \frac{\Delta a}{a_0} \times 100 \text{ per cent}$$

It is generally considered that for effective nucleation a misfit of less than 10 per cent is desirable: some authorities consider this figure to be excessive, and others permit values up to 15 per cent. It is also generally

accepted that the misfit should be calculated on close-packed planes of the crystals under consideration, and that the lattice correspondence should be two-dimensional.

An example of a two-dimensional lattice correspondence is shown in Fig. 13 where projections of (100) planes of two substances have been superimposed.

Cibula¹⁴ has used these principles extensively in a study of grain refinement of non-ferrous alloys, and Dennison and Tull¹² utilized the theory with considerable success to select an inoculant for Cu-Al alloys containing the beta phase. In the present research, a survey of crystal structures has been made in an attempt to find suitable inoculants for the delta phase in high-speed steel.

It is considered necessary to inoculate the primary phase rather than endeavour to induce a modification of the eutectic colonies. Fig. 14 shows the structure of the eutectic normally found in high-speed steels, and

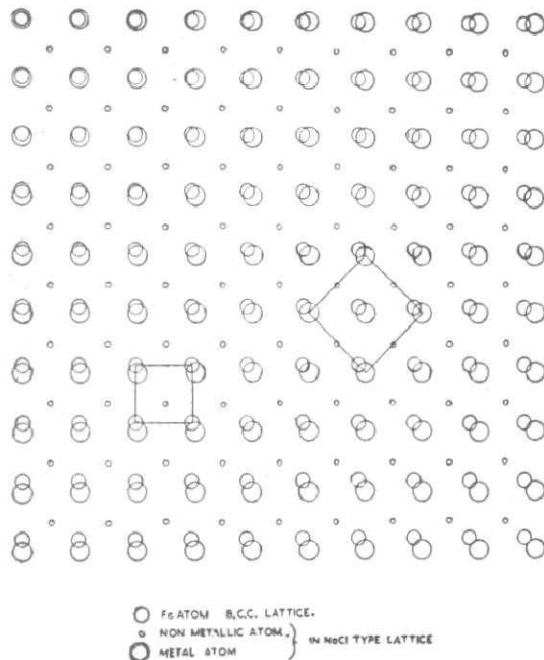


FIG. 13 — CORRESPONDENCE OF LATTICES OF DIFFERENT SUBSTANCES

Fig. 15 shows a structure in which some alteration of the normal habit has taken place. Here, although the actual eutectic structure has been broken, it is still apparent that the carbide exists in large colonies.

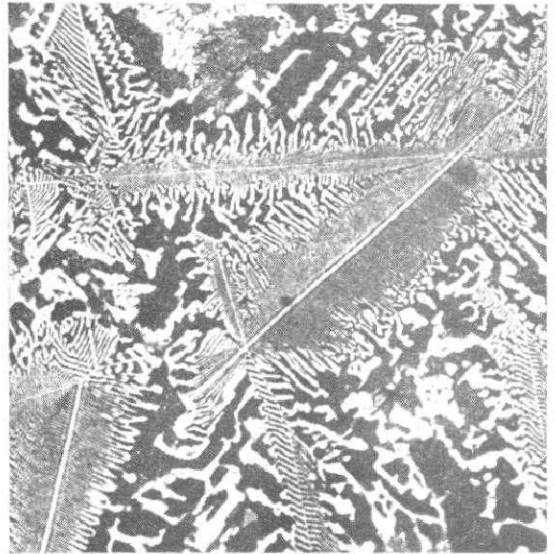


FIG. 14 — NORMAL EUTECTIC IN HIGH-SPEED STEEL.
 × 100

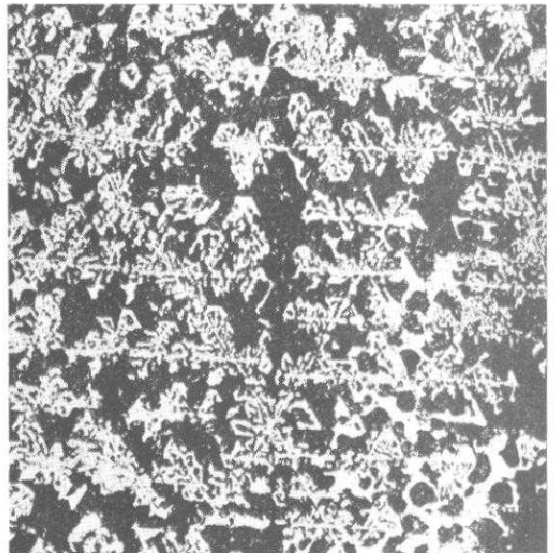


FIG. 15 — MODIFIED EUTECTIC IN HIGH-SPEED STEEL.
 × 100

Effects of Vibration

Mechanical vibrations are known to have an important disturbing influence on liquid \rightarrow crystal reactions¹⁷, and quite mild vibrations can induce nucleation in undercooled liquids at temperatures higher than those at which solidification would occur under conditions free from vibration. One theory¹⁸ suggests that the effect is caused by cavitation, at the container wall or elsewhere, and that waves of high alternating pressure are formed by the rushing of liquid into these cavities. A further theory suggests that vibration may cause disruption of crystals already present and thus give rise to an increased number of nuclei.

Support for this is found in the field of non-ferrous metallurgy and various experiments have been carried out by manufacturers to explore the effect of vibration on the solidification of high-speed steel ingots. No success has been publicly reported, but one interesting result is shown in Figs. 16 and 17. These photographs show the effect of vibration at 50 cycles per second on the structure of an ingot of high-speed steel¹⁹. The cause of the banding effect is not known, but it has been suggested that it may be due to a resonance phenomenon.

Vibration at ultrasonic frequencies has also been used with some success in metallurgy.

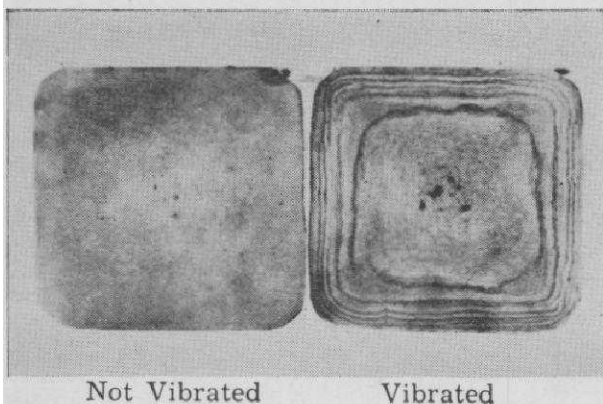


FIG. 16 — COMPARISON OF HIGH-SPEED STEEL INGOTS VIBRATED AND NOT VIBRATED DURING SOLIDIFICATION

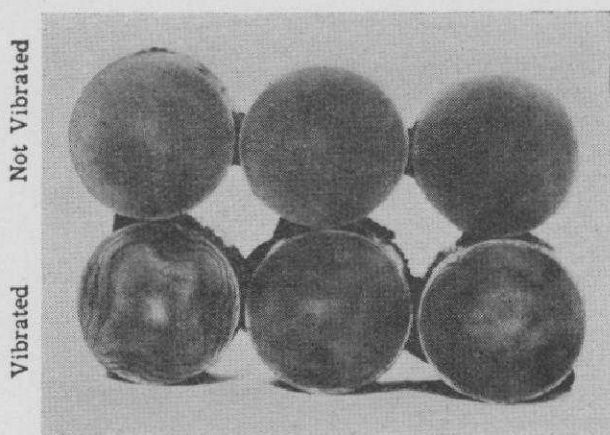


FIG. 17 — BILLETS PRODUCED FROM FIG. 16

(Figs. 16 and 17 are reproduced by kind permission of Mr. J. Woolman, Brown-Firth Research Laboratories)

Alder²⁰ reports an experiment in which the grains in a zinc ingot have been refined in this way. The chief difficulty in this method appears to be that of transferring the energy from the source of vibration to the melt, and a well-known firm of electronic instrument manufacturers has expressed the view that the method cannot as yet be applied economically to any but the smallest ingots of high-speed steel.

Electromagnetic Stirring during Solidification

This method has been tried in an attempt to preserve uniformity in the liquid phase during solidification and to assist the freezing process by bringing the reactants together by mechanical means instead of relying on the slow diffusion processes normally involved. On the laboratory scale this has been achieved by allowing a melt to solidify in the high-frequency furnace with a power input sufficient to stir the metal and yet sufficiently low to allow cooling to take place. Very slow cooling was obtained and the structures were reasonably fine, although there was inevitably eutectic present and the normal coring was observed.

Conclusion

The difficulties encountered in this type of work are manifold: for instance, the high cost of the steel makes research work expensive, and manufacturers are understandably loath to risk the possible wastage of an ingot weighing, say, half a ton, which may be worth something in the region of £250. Other difficulties are attributable to the long annealing and other heat treatment operations required when dealing with high-speed steel. Nevertheless, manufacturers are alive to the problems and they, as well as B.I.S.R.A., are actively engaged in exploring new methods to endeavour to improve still further the high quality of the tool steels produced in Sheffield — for it is here that practically all the high-speed steel made in Britain is produced.

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