

Fig. 2. Three dimensional representation of flake graphite as it develops during eutectic solidification.

iron (where graphite assumes a spheroidal shape). In both cases solidification of the liquid may entail both the stable austenite-graphite eutectic as well as the metastable austenite-carbide eutectic.

Flake Graphite Formation

Solidification of the austenite-flake graphite eutectic occurs by the nucleation of a eutectic cell that grows with both phases in contact with the liquid, Figure 1. Graphite appears to form a number of independent flakes within the cell, but in fact all of the flake graphite within a cell is interconnected and has developed as a result of the branching of flakes in order to optimize the interflake diffusion distance for a given localized solidification cooling rate, Figure 2. Within the flake the graphite crystal structure is well defined, with the basal planes of the hexagonal crystal aligned along

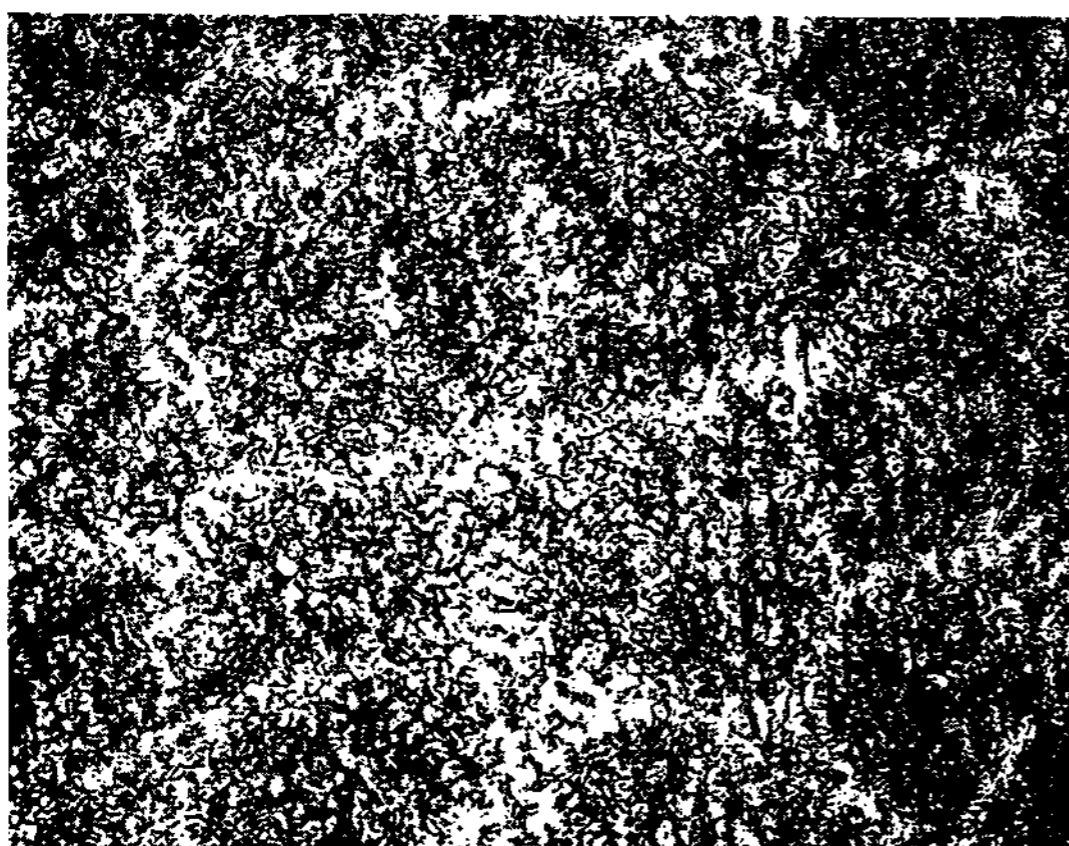


Fig. 3. Typical cell structure in gray cast iron revealed by etching with Stead's reagent. This technique deposits copper on the polished surface except in those areas which are high in phosphorus content.

the length of the flake. Curvature of the flake (induced by the co-precipitation of austenite) causes low angle grain boundaries to form within the flake, and branching is associated with more significant crystallographic defects. Growth of the flake occurs by the addition of carbon atoms from the liquid onto the exposed ends of graphite basal planes.

It would be expected that in a complex alloy such as commercial gray cast iron, segregation of certain elements would accompany the solidification process. Some elements (e.g., Ni, Si, etc.) segregate negatively with higher concentrations near the origin of the eutectic cell, while others (e.g., Mn, Cr, P, etc.) segregate positively with higher concentrations in the regions where growing cells impinge upon each other (the last regions to solidify). An example of this is shown in Figure 3 where Stead's reagent has been used to delineate those intercellular regions of higher phosphorus content.

Since diffusion controls the interflake distance in the eutectic formation it follows that the flake morphology will depend upon nucleation and growth temperatures encountered during solidification. If minimal undercooling is realized the austenite-flake graphite eutectic is nucleated at the highest temperature in the eutectic temperature range. Typically the heat evolved is

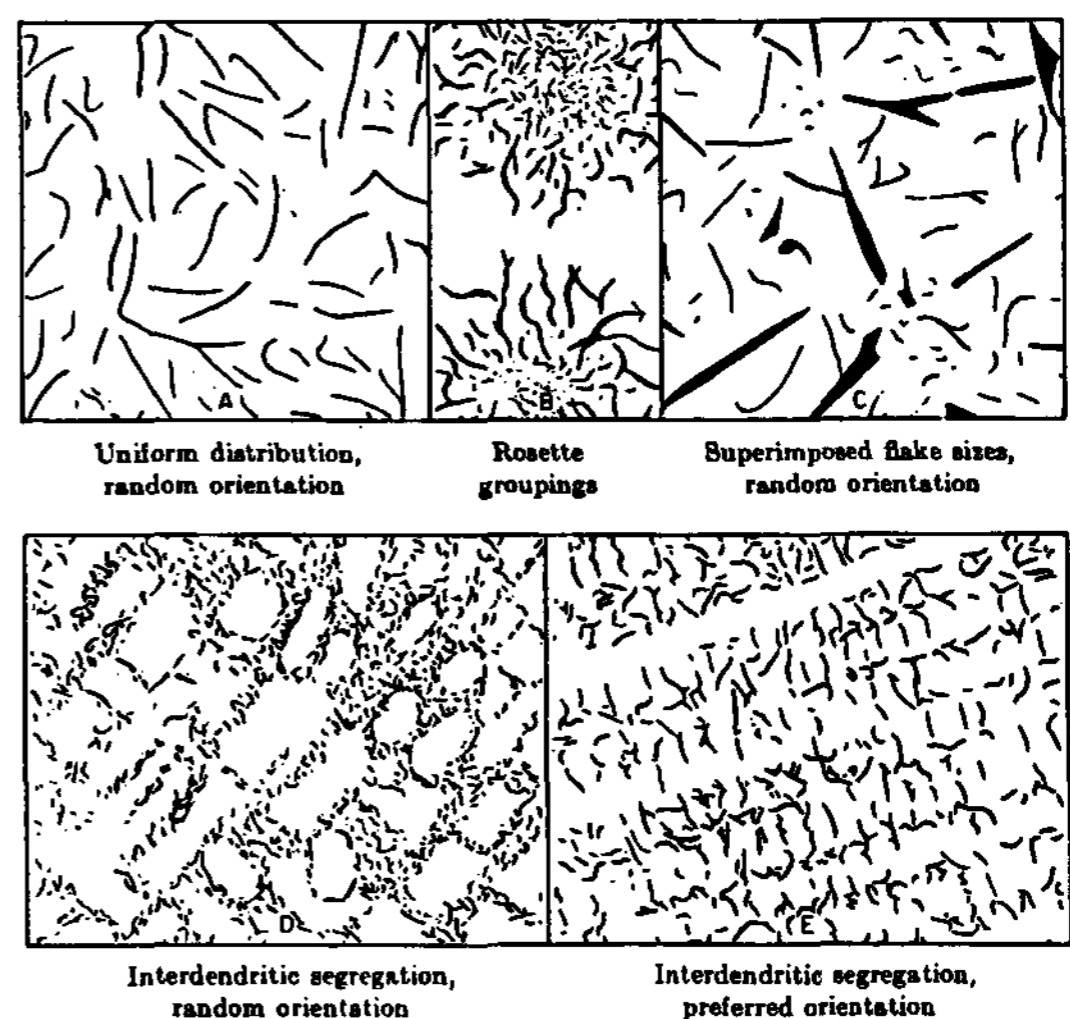


Fig. 4. Graphite classification as presented by the American Society for Testing and Materials (ASTM), ASTM A247.

sufficient to maintain this temperature throughout the bulk of eutectic solidification and the graphite develops as ASTM type A (randomly oriented, uniformly distributed flakes), Figure 4. Undercooling due to inadequate eutectic nucleation and/or more rapid solidification cooling rates results in nucleation of the eutectic cell with a smaller interflake distance, but when the heat of fusion raises the local solidification temperature (recalescence) the interflake distance coarsens resulting in ASTM type B graphite (rosette shaped cells). Further undercooling is accompanied by the formation of well developed austenite dendrites prior to nucleation and growth of the eutectic which is restricted to the interdendritic regions. Austenite-flake graphite cells nucleate and grow within these interdendritic regions with small interflake distances (due to the undercooling). The resulting eutectic structure is termed ASTM type D graphite (randomly oriented finely dispersed flake graphite restricted to interdendritic regions). With further undercooling diffusion of carbon is inhibited to such a degree that graphite no longer forms from the liquid, rather the metastable austenite-carbide eutectic develops.

This solidification "ladder" is therefore the result of undercooling due to inadequate nucleation and/or increased solidification cooling rates. For example, if a given iron is poured into a wedge shaped casting of suitable dimensions the iron carbide eutectic will form at the tip of the wedge (high surface area to volume ratio). This structure will gradually change to ASTM type D graphite, ASTM type B graphite and ASTM type A graphite with increasing distances from the tip of the wedge due to the slower solidification cooling rates encountered. Effective nucleation, higher carbon and/or silicon contents, graphitic charge materials, proper melting procedures, etc. result in nucleation of the eutectic at high temperatures and the formation of ASTM type A graphite¹.

Spheroidal Graphite

In 1948 the development of a procedure whereby

¹ASTM type E graphite is identical to ASTM type D graphite except that when the austenite dendrite arm thickness is increased (e.g., lower carbon equivalent) eutectic solidification is more severely restricted resulting in fine graphite flakes, preferentially aligned within the interdendritic regions.

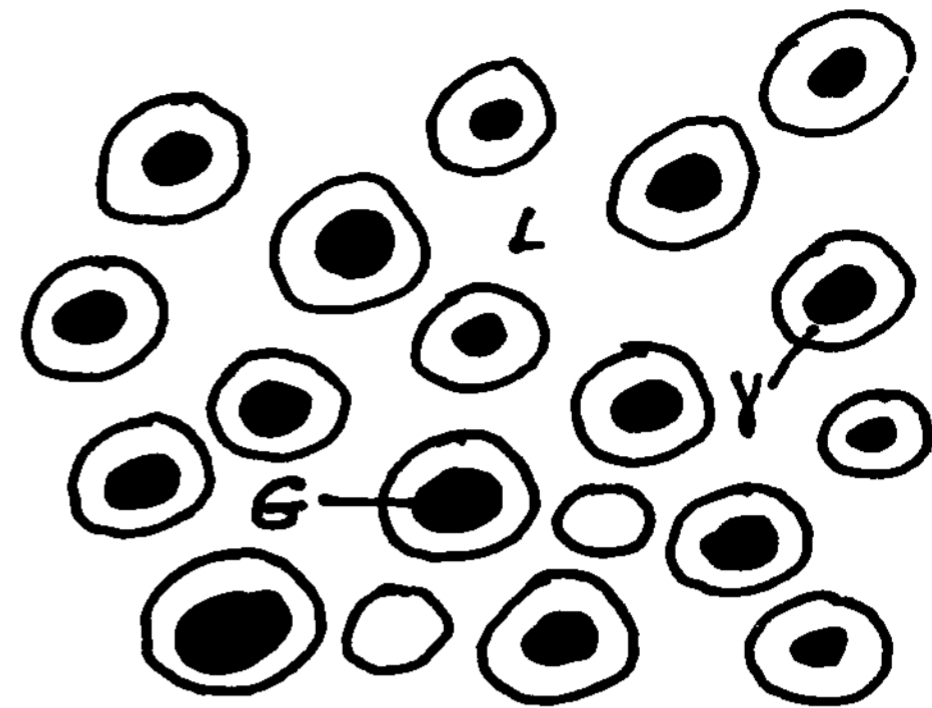


Fig. 5. Sketch illustrating the development of the spheroidal graphite-austenite cell during eutectic solidification.

spheroidal graphite could be formed during cast iron eutectic solidification was announced. Ductile iron (also called, nodular iron, spheroidal graphite cast iron, etc.) formed from the addition of a small amount of a nodularizing elements (e.g., Mg, Ca, Na, Ce, etc.) to a low sulfur content molten cast iron. In this case eutectic nucleation resulted in the formation of a cell wherein graphite grew in a spheroidal form largely with the graphite enveloped in a shell of austenite, Figure 5. Since most commercial ductile iron compositions are hyper-eutectic, the graphite spheroid typically develops as a pro-eutectic phase and later becomes surrounded by an austenite shell. The unit cell becomes more complex, however, as shells impinge upon one another and as austenite dendrites develop due to solidification over a temperature range (longer than for gray cast iron, with changing composition of the austenite). More effective nucleation resulted in increased numbers of eutectic cells and more widely dispersed segregation of the last liquid to solidify. Lower cell counts (nodule counts) result in solidification of the eutectic liquid over a wider temperature range (solidification being restricted by diffusion of carbon through the austenite shell surrounding the graphite spheroid) with the result that eutectic carbides can develop in the intercellular (inter spheroidal) regions. As expected, increased solidification cooling rates will be accompanied by an increase in the number of eutectic cells (spheroids). However, increased cooling rates will also result in the formation of the metastable austenite-iron carbide eutectic. In fact, the effective nodularizing elements are typically elements which not only pro-

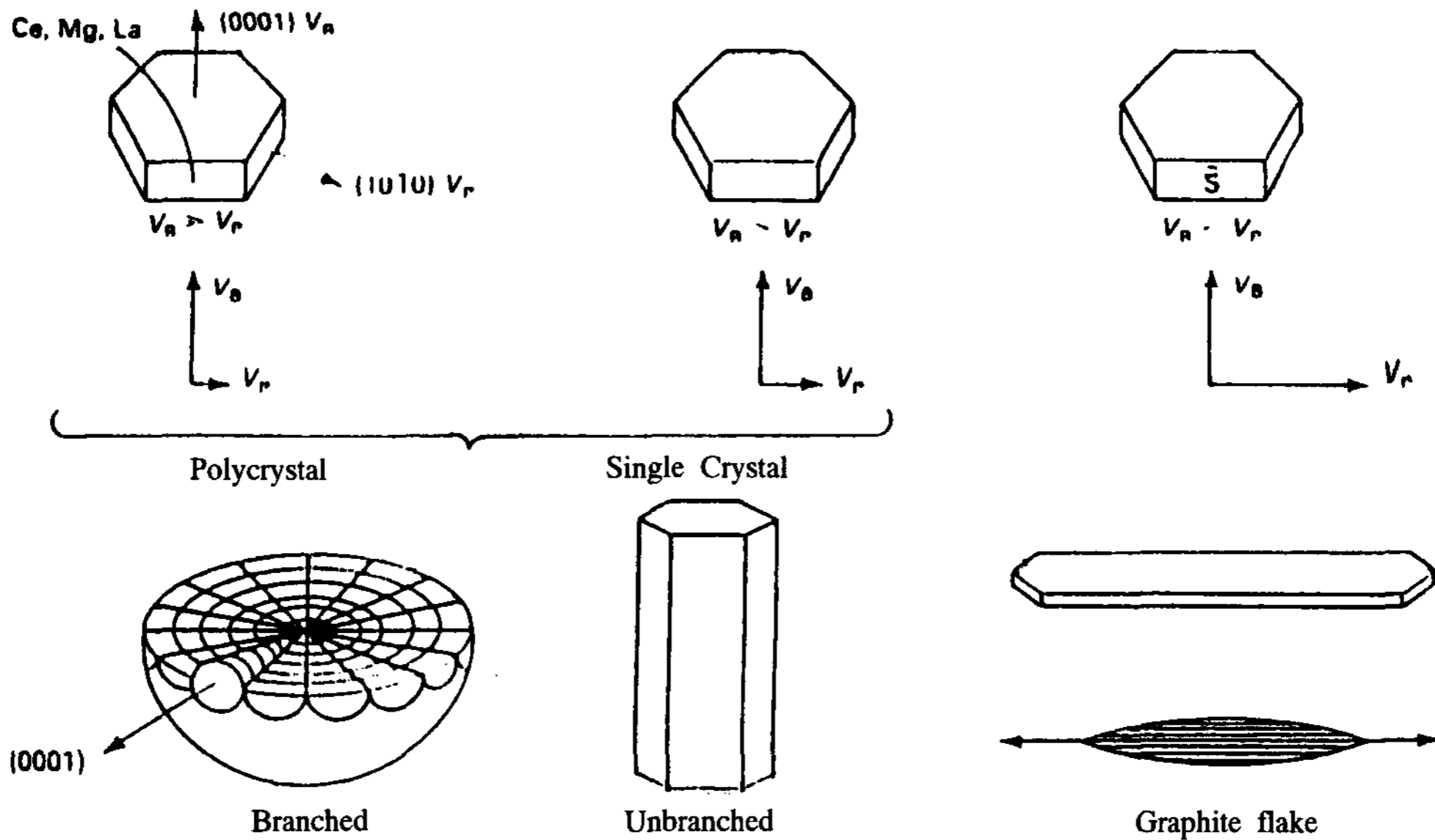


Fig. 6. Crystallography of graphite.

mote carbide formation, but increase the temperature of eutectic carbide formation.

Non-spheroidal or deleterious graphite morphologies in ductile irons can be attributed to the interference of spheroidal graphite formation by the presence of other elements (e.g., vermicular graphite due to Ti, or low effective Mg content; chunky graphite due to Ce or rare earths; spiky graphite due to Pb; etc.).

Graphite Crystal Structure

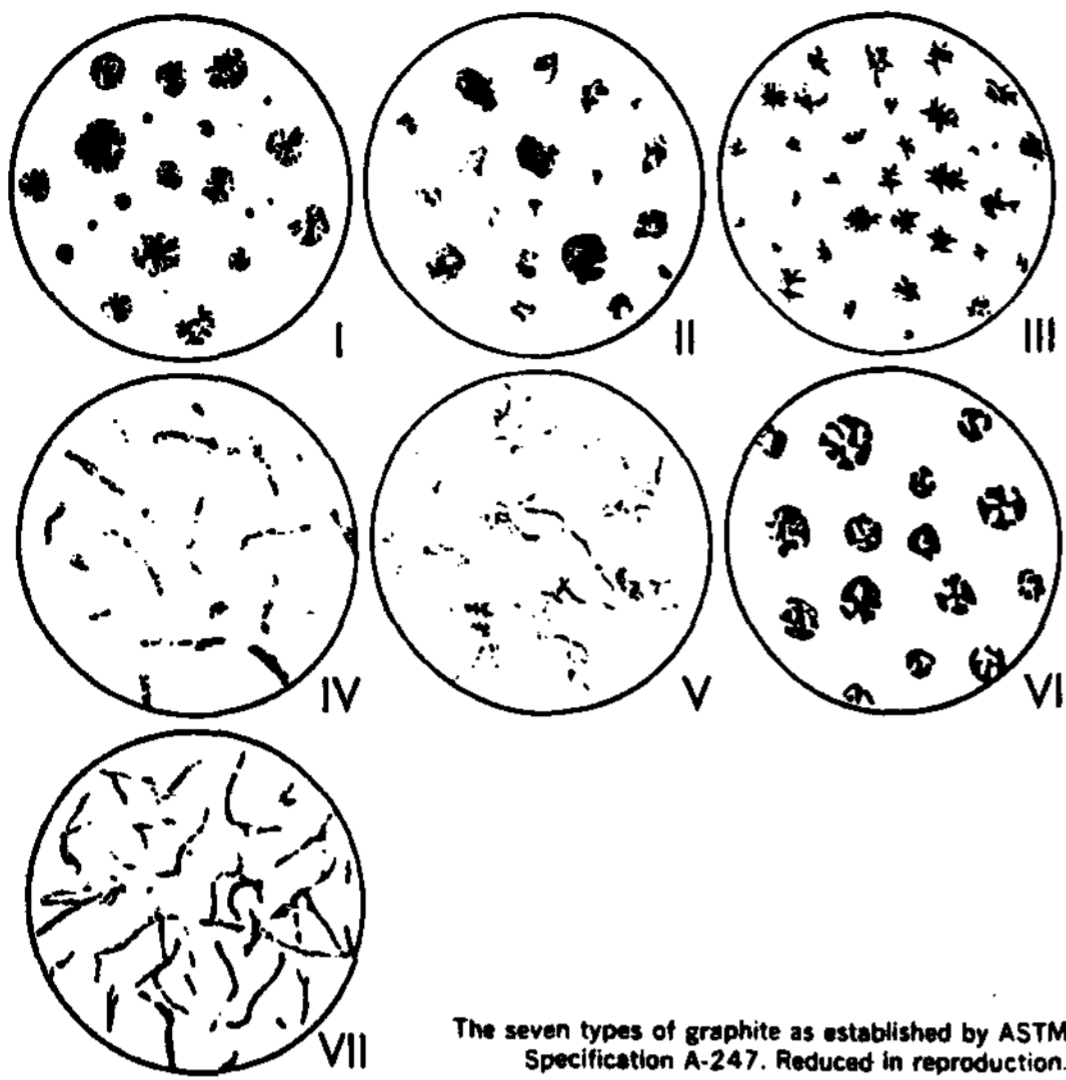
Graphite is an hexagonal crystal structure bound by six prism faces (a faces) and two close packed hexagonal basal planes (b faces), Figure 6. Graphite has a high entropy of fusion, so that growth occurs along the pole of the plane with the lowest interfacial energy in contact with the melt. In the absence of surface active elements carbon atoms are preferentially incorporated into the basal planes. It is of interest to note that while the apparent growth of spheroidal graphite is in the c direction (normal to the basal plane) the actual growth direction of the crystal is in the a direction (normal to the prism plane) for both spheroidal and flake graphite.

The crystallographic structure of flake graphite and of spheroidal graphite differ in that the basal planes of the graphite are parallel to the side walls of the flake shapes. During solidification (with frequent low angle

grain boundary development, as well as twinning), the basal planes are exposed to the liquid, so that carbon atoms readily deposit on the edges of the graphite basal planes. However, in ductile iron the graphite basal planes lie parallel to the spheroid surface, the result of a myriad of low angle grain boundaries developed as the spheroid increases in diameter. Growth of the spheroid occurs by the addition of carbon atoms onto the edges of the basal planes, sometimes described as being similar to the growth of a cabbage leaf. Growth is dependant upon the existence of a screw dislocation within each segment of the graphite spheroid.

The development of graphite as a flake or spheroidal shape depends upon growth restrictions during solidification. Treatment of molten iron with a sufficient amount of nodularizer insures the development of spheroidal graphite. But if an insufficient amount of nodularizer is present, growth will take place in a complex, intermediate manner (such as the formation of vermicular, or compacted, graphite).

For flake graphite hyper-eutectic irons the graphite is nucleated in the melt and can grow in a relatively unrestricted manner developing plates which are straight, but often branched (type C graphite), Figure 4. Since the graphite phase is considerably less dense than is the liquid, these plates will float to the melt surface



The seven types of graphite as established by ASTM Specification A-247. Reduced in reproduction.

Fig. 7. ASTM classification of graphite morphologies in cast irons. Types I and II are generally acceptable for ductile iron production. Type III is actually the form of graphite obtained in malleable irons which are always heat treated. Types IV, V and VI are undesirable graphite forms: vermicular, or compacted, graphite (type IV), crab like graphite (type V), and exploded graphite (type VI). Type VII is flake graphite, refer to Figure 4.

when they reach a size determinable by Stoke's Law. This flotation process effectively removes carbon from the bulk melt and the melt composition may gradually approach the eutectic. Graphite may also be nucleated at the melt surface, i.e., along the side walls or at the top surface of the melt. When this occurs, that pro-eutectic graphite at the surface develops as large platelets of graphite with the graphite basal plane extended along the platelet, a product referred to as kish.

Hyper-eutectic spheroidal graphite is also subject to flotation when the size of the graphite spheroid becomes large enough to float according to Stoke's Law. This mobility of the graphite exposes the graphite spheroid to a variation in local melt composition and temperature, causing irregularities to develop in the crystal structure at the graphite surface. These irregularities, combined with the co-precipitation of austenite at lower melt temperatures results in the formation of exploded graphite, Figure 7, a not uncommon occurrence in highly hyper-eutectic composition castings which solidify slowly. (It should also

be noted that this destabilization of the spheroidal graphite growth interface occurs as a result of melt stirring, either intentional or induced naturally due to changes encountered in localized melt density during solidification.) Flotation of spheroidal graphite is more readily encountered at low nodule counts (less effective nucleation) since the average diameter of the spheroid will be larger for a given volume fraction graphite.

Once the temperature of the melt is low enough for the eutectic to solidify, it will develop an austenite shell around the spheroid increasing its density and reducing the tendency for flotation to occur. As noted previously, the eutectic must now form by the diffusion of graphite through the austenite shell, with growth of the austenite spheroid along with increased thickening of that surrounding austenite shell.

Nucleation of the Eutectic

With this introduction, the subject of nucleation of the eutectic will be considered. Through trial and error experiments, foundrymen have concluded properly that effective nucleation of the eutectic in cast iron melts can be achieved by the addition to the melt of small amounts of graphite and/or ferro silicon. Both materials have a role in the production of commercial cast iron when formulated and used properly.

Nucleation of the Eutectic With Ferro Silicon

The role of ferro silicon in the nucleation of graphite has been studied extensively.

However, these studies have naturally concentrated on the nucleation events associated with spheroidal graphite formation since it is relatively easy to identify the non-graphite substrate that is always observed at the growth center of graphite spheroids. On the other hand, it is quite difficult to locate the growth center of a flake graphite cell much less identify the characteristics of the substrate present there.

Nevertheless, a number of theories have been proposed in an attempt to explain the relationship of a nucleating substrate to graphite, e.g., silicon carbide theory, salt like carbide theory, sulfide/oxide theories, etc. these theories are all based upon the assumption that

graphite is nucleated as a heterogeneous event occurring during solidification, and that the presence of certain minor elements are important in the nucleation process, i.e, Ca, Al, Zr, Ba, Sr, Ti, etc. A compound is formed that serves as the substrate for graphite nucleation by providing a crystal structure with minimal crystallographic disregistry (similar to that observed for coherent/semi-coherent interfaces) which lowers the surface energy of the system.

Role of Compound (Inclusion) Formation in Cast Irons Melts. It is feasible for a variety of compounds to form within a cast iron melt: sulfides, oxides, nitrides, silicates, etc. Sulfides and oxides of the Group IIA elements (Mg, Ca, Sr, Ba) form face centered cubic crystals of the NaCl type, have similar lattice parameters and can be matched with those of graphite, and have high melting points (so that they are solid phases during cast iron solidification). Silicates such as $XO \cdot SiO_2$ and $XO \cdot Al_2O_3 \cdot SiO_2$ (where X may be Ca, Ba, Sr, etc.) have similar lattice parameters and a high melting point. These compounds, then, could be candidates for heterogeneous nucleation sites based upon lattice disregistry theory.

It is recognized that the nucleus/graphite interfacial energy is the controlling factor in heterogeneous nucleation behavior. But reliable data of this type is not available at this time, so that evaluation of the effectiveness of various substrate candidates for nucleating graphite must be based upon the planar lattice disregistry model alone.

Studies made of the inclusions present in spheroidal graphite irons have clearly demonstrated that the composition of those inclusions is directly related to the composition of the alloy used for melt treatment (nodularization) or for inoculation.

However, three categories of inclusions have been proposed:

Type A-Inclusions containing Mg, Ca, S and Si which have been identified both at the spheroidal graphite growth center and in the matrix, and which are angular or faceted in shape.

Type B-Inclusions containing Mg and Si, which are found in the matrix and are typically rectangular in shape.

Type C-Inclusions containing Mg, P and rare earths,

which are located primarily in the interdendritic regions and which are large, irregular shapes.

It has been suggested that the Type A inclusions are effective as substrates for graphite nucleation. These type A inclusions have also been found to have complex, duplex structures. For example inclusions studied after magnesium treatment of the melt using a proprietary ferro silicon alloy were observed to have a sulfide core (Mg, Ca, S) with an outer shell devoid of sulfur (Mg, Si) and proposed to be a magnesium silicate or forsterite. After subsequent inoculation these type A inclusions contained calcium and aluminum at the surface, proposed to be anorthite.

High Purity Ferro Silicon Alloys. It is also well recognized that the use of high purity ferro silicon alloys (regardless of the percentage of silicon present) is ineffective in obtaining an inoculation effect in either gray or ductile cast irons. Effective silicon containing inoculants must contain the elements previously mentioned in a form where they can react within the melt to form suitable substrates for the nucleation of graphite, i.e., they are not effective when present in, or with, the ferro silicon alloy as oxides.

Fading Effect. The well known fading effect of inoculants in cast iron melts has been described as due to the agglomeration and coarsening, Ostwald ripening, of inclusions with time-a practice that greatly reduces the potential number of substrate particles and is observed as decreasing nodule count in spheroidal graphite iron

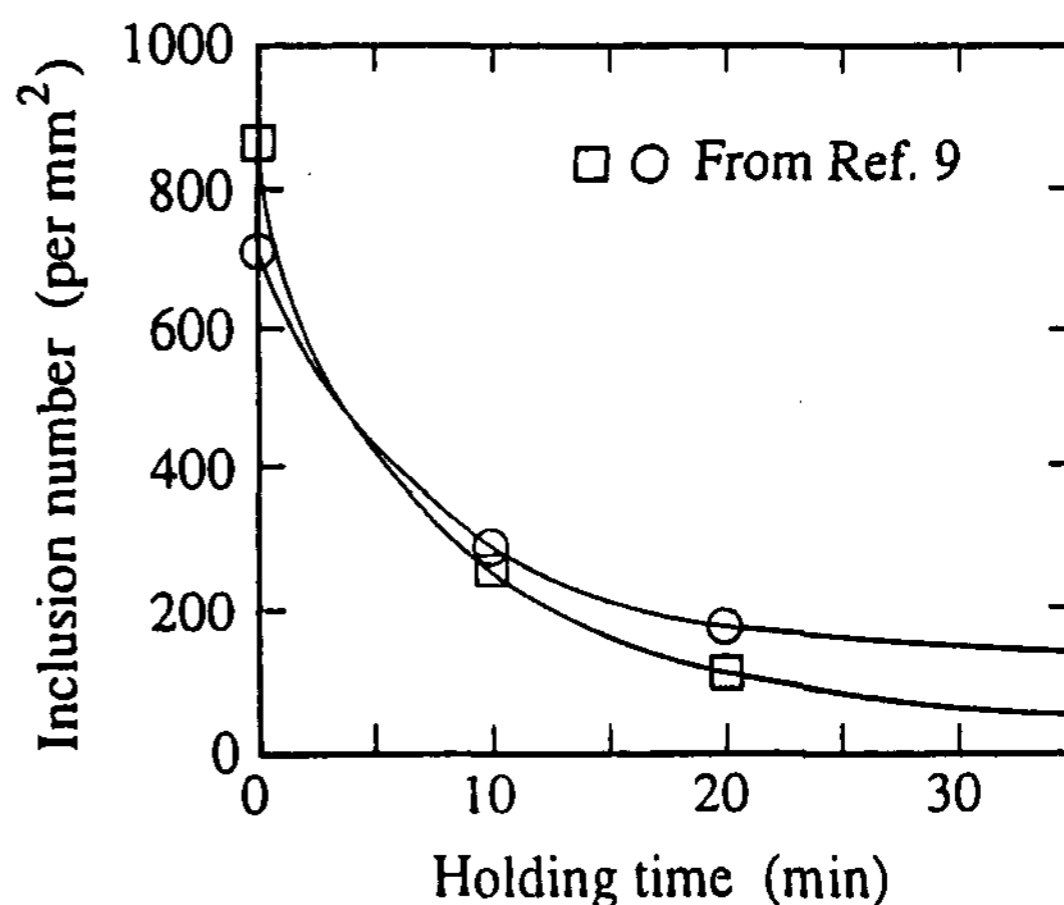


Fig. 8. Inclusion count reduction with time as determined by calculation and by experiment.

and decreased cell count in flake graphite iron, Figure 8. This inoculation fading effect, whether associated with the magnesium treatment of a melt or when inoculating agents are added, must be separated from the fading of magnesium in the production of ductile or compacted graphite cast irons. While magnesium may participate in the formation of substrates suitable for graphite nucleation, the role of magnesium in the treatment process is primarily one of controlling the growth conditions of graphite (e.g., flake vs. compacted or vermicular vs. Spheroidal graphite).

Summary. Nucleation from the addition of ferro silicon can then be summarized as being a process which involves the formation of suitable inclusions within the cast iron melt, where those inclusions present a crystal structure which effectively matches that of graphite. Fading of the inoculation effect is attributed primarily to the coarsening of these inclusions occurring over relatively short periods of time.

The cast iron foundryman should be well aware of the fact that in order for a ferro silicon inoculant to be effective it must contain those elements which are necessary to form compounds in the melt which act as substrates for the nucleation of graphite. Foundry grade ferro silicon alloys, including both commercial foundry grade ferro silicon and proprietary ferro silicon, are formulated to contain amounts of Ca, Ba, Mg, Zr, Al, etc. that will insure their effectiveness as cast iron inoculants, and must be distinguished from commodity ferro silicon which is used principally as a source of silicon in alloying, for deoxidation, etc. These effects are temperature and time dependant, with fading taking place more rapidly at higher melt temperatures but where fading is always an important metallurgical con-

trol, factor. For example, significant loss of magnesium treatment or inoculation is observed in time periods as short as 8 minutes.

Nucleation of the Eutectic With Graphite

Graphite was used by cast iron foundrymen as an inoculant prior to the introduction of ferro silicon alloys for that purpose. However, the mechanism whereby graphite was effective in this respect was never well understood in spite of the fact that graphite was obviously a perfect substrate for graphite growth so that nucleation of graphite was never a consideration. In this way graphite is highly effective in causing the cast iron eutectic to form at the highest possible temperatures.²

Concerns Regarding Graphitic Inoculants. The use of graphite in cast iron production, however, presents certain difficulties which must be understood if they are to be overcome.

a. Graphite must be effectively distributed in the molten iron. In order for this to occur the graphite addition must be wetted by the molten iron, a fact that is accomplished with relative ease in gray iron melts, but which is difficult, or impossible, in ductile iron melts due to the influence of magnesium on the surface tension of the melt with respect to graphite.³

b. Once the graphite is effectively wetted by the melt, the graphite must be distributed through the melt as solid particles upon which the eutectic may form. The mechanism whereby this distribution best occurs is not yet well understood.

c. The effectiveness of graphite as an ideal substrate for eutectic solidification depends upon the existence of a solid particle in the melt (or perhaps a liquid cluster that will form that initial solid particle of graphite). However, when graphite is added to the melt it will enter into solution in the melt at a rate dependant upon melt temperature and holding time. Once in solution it will no longer be available as a substrate for eutectic

²It must be recognized that carbon additives are not necessarily graphitic, and that the crystallinity of various carbon additives may vary significantly. For example, materials having a crystallinity under 100 Å are not effective as inoculants, and it appears that when the crystallinity exceeds about 1000 Å the effectiveness is less. Peak effectiveness appears to be obtained at about 300 Å. Examples of materials that may be encountered in cast iron operations are: a) Metallurgical coke, $L_c < 20 \text{ Å}$; b) Petroleum coke, L_c about 90 Å; c) Purified Mexican graphite, L_c about 200 Å; d) High purity graphite, L_c about 300 Å; e) Electrode scrap graphite, L_c about 500 Å; f) Crystalline graphite, L_c about 700 Å; g) Natural graphite (Sri Lanka), $L_c > 1000 \text{ Å}$; h) Natural graphite (China), L_c about 2000 Å.

³The principle characteristics of the molten metal which affect the rate of carbon pick up and the efficiency of carbon recovery may be summarized as follows: 1) Melt chemistry and temperature; 2) Dissolution kinetics; 3) Sulfur content of the melt; 4) Relative agitation of the melt; 5) Physical condition of the melt, including prior magnesium treatment of the melt.

growth.

Nevertheless, inoculation of gray cast irons with graphite has a long history in cast iron production, and is utilized particularly in the production of thin section castings, or castings which have a high tendency to develop chill. Its use must be carefully controlled in order to prevent excessive inoculation, and it is often used in conjunction with foundry grade ferro silicon or proprietary ferro silicon alloys in critical applications. In ductile iron production graphite is effective as a melt pre treatment, being added prior to, or in conjunction with, magnesium treatment of the melt. However, it is not effective when added as a conventional post inoculant, and its use in this manner will result in reduced nodularity.

Mechanics of Graphite Nucleation. In order to better understand the role of graphite in the solidification of cast irons certain experiments have recently been conducted using cast iron melts containing minor element chemistry characteristic of commercial melts. These experiments utilized the droplet emulsion technique, a procedure which has been used to study a number of alloy systems but has not yet been applied to ferrous melts.

The relatively high melting temperatures encountered in the melting of cast irons, as well as the reactivity of these melts with the environment, demanded that this technique be modified somewhat. It would be necessary to develop a carrier material that could be used over the range of temperatures encountered, and that would provide sufficient protection of the droplets from oxidation, decarburization and/or silicon loss. The carrier solution used consisted of a slag composed of 70% SiO₂-20% Al₂O₃-10% CaO. This slag protected the emulsions against agglomeration, and provided an essentially neutral environment for the droplets. However, where strongly hypo-eutectic alloys were being studied the aforementioned slag was treated with 15% FeO, and for hyper-eutectic alloys the slag was treated with 15% C, to insure minimum reactivity with the droplets. After heating and cooling cycles were applied during DAT measurement these slags were glassy, but contained some crystalline phases a condition preferred in terms of promoting the largest level of liquid undercooling.

The alloy to be studied was melted and rapidly solidified by melt spinning in order to produce a thin ribbon which was crushed to yield flake shapes of white iron that were separated into five size ranges: <20, 20-44, 44-62, 62-89 and >89 μm. It was recognized from previous studies that the undercooling behavior of droplets is dependant upon the droplet size. Droplet microstructures exhibited a uniform microstructure. A mixture was then prepared of powdered slag and white iron flakes which was heated in an Al₂O₃ crucible. The white iron flakes melted first, and then the powdered slag, yielding a molten slag containing molten metal droplets. On cooling the solidified slag contained isolated metal droplets which provided for nucleant isolation and droplet independence. The sequence of steps involved in the preparation of the droplets is summarized in Figure 9.

Undercooling and solidification behavior was studied using a Perkins-Elmer DTA 7 system in which the mixture of white iron flakes (20-44 μm) and slag was heated and cooled in an alumina crucible at 20°C/min. the metal:slag ratio used in these experiments ranged from 7:3 to 6:4, values that were found to prevent agglomeration of the droplets.

Two types of experiments were conducted using this technique. The first is referred to as a "homogeneous cycle", involving heating the droplets to a temperature above their liquidus temperature, followed by cooling to lower temperatures. The second technique is referred to as a "heterogeneous cycle" where heating was limited to a temperature between the eutectic temperature and the liquidus temperature, followed by cooling. In the latter case the primary phase (austenite or

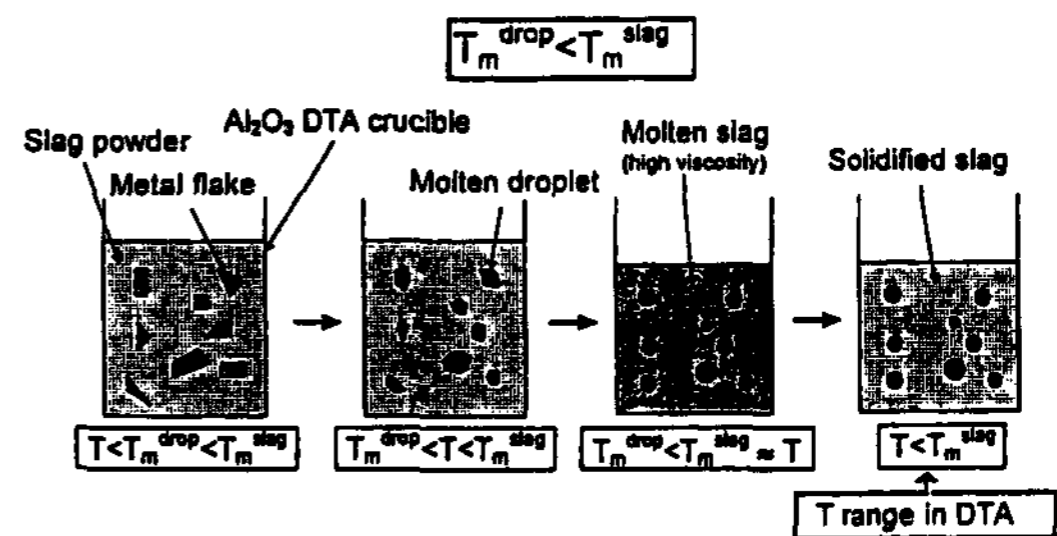


Fig. 9. Sketch illustrating the steps involved in the preparation of the droplet emulsion technique reported in this work.

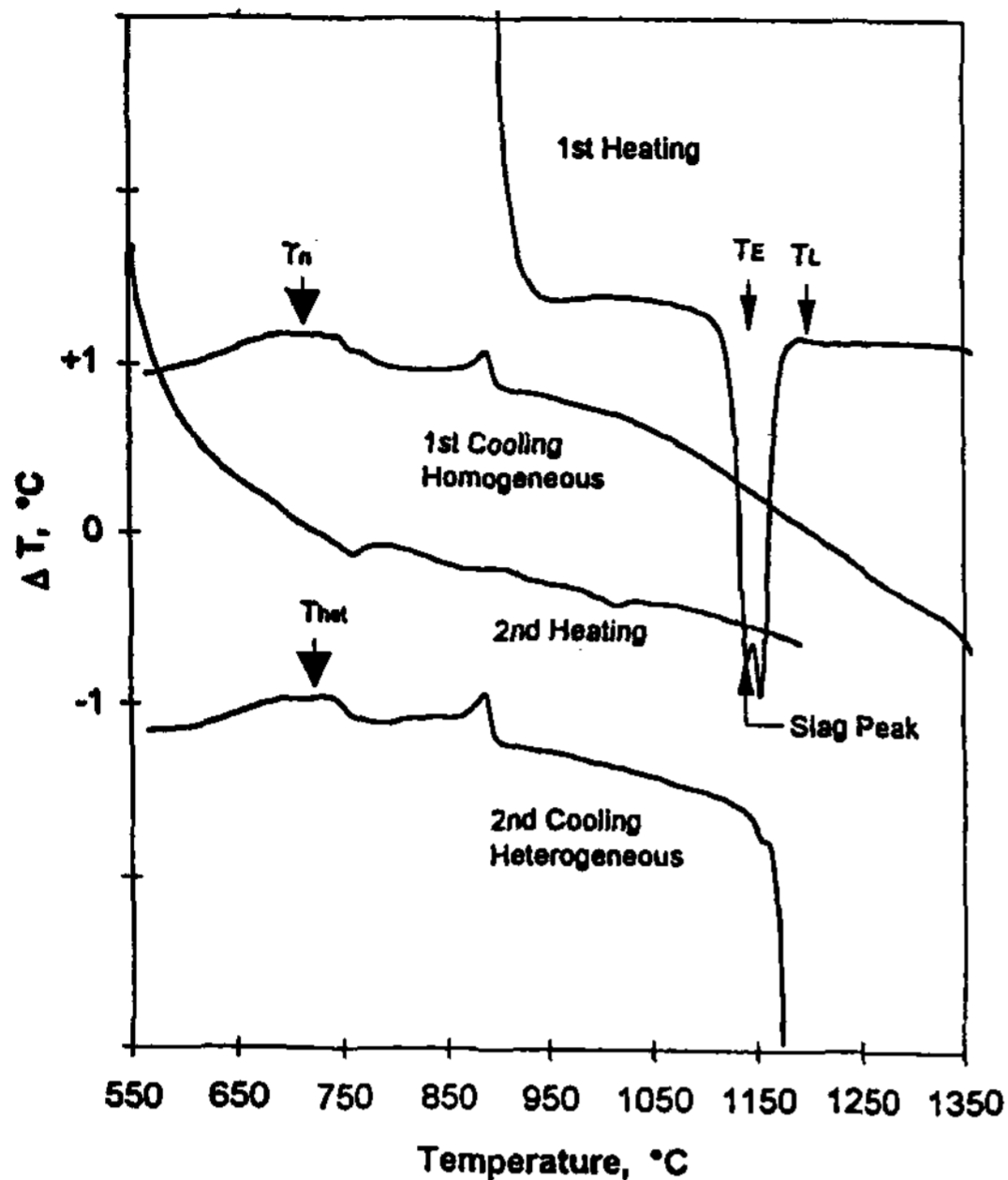


Fig. 10. Typical DTA heating and cooling curves obtained from a hypoeutectic droplet emulsion sample of cast iron. After subjecting the sample to the "homogeneous" cycle it was then subjected to the "heterogeneous" cycle.

graphite) was present in the droplets prior to their being cooled. DAT heating and cooling curves for the "homogeneous" and "heterogeneous" cycles are illustrated for a typical droplet emulsion in Figure 10.

Three alloys of approximately 1.91% Si were studied at carbon contents of 2.86% (hypo-eutectic), 3.45% C (slightly hypo-eutectic) and 4.14% C (hyper-eutectic). The undercoolings achieved with both the "homogeneous cycle" and the "heterogeneous cycle" are plotted on a 2.5% Si isoplethal section of the Fe-C-Si phase diagram, Figure 11.

It may be observed that under conditions of the "homogeneous cycle" significant undercooling were reported. For the 2.86% C, 1.91% Si alloy the maximum undercooling was about 620°C the largest reported undercooling of Fe base alloys!). This alloy solidified as a single phase carbon supersaturated metastable austenite which subsequently precipitated a fine dispersion of carbide particles in amounts corresponding to the alloy composition (2.86% C). The 3.45% C alloy subjected to the same thermal cycling resulted in an undercooling of about 520°C and a single phase carbon

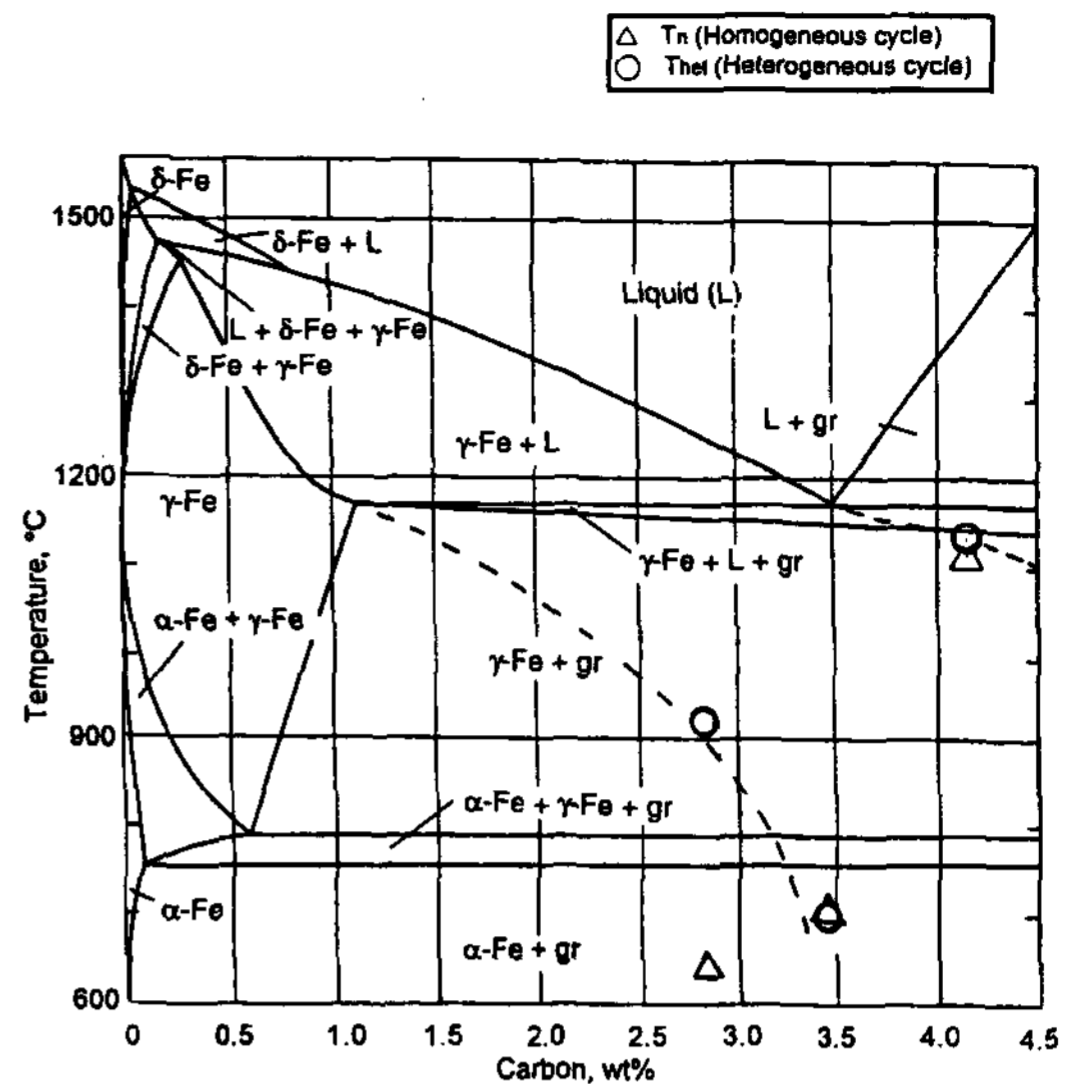


Fig. 11. Summary of the undercooling results plotted on a 2.5% Si isoplethal section of the Fe-C-Si phase diagram. Results obtained from both "homogeneous" and "heterogeneous" cycles are illustrated.

supersaturated metastable austenite, Figure 12, an austenite having a carbon content of 3.45%! The hyper-eutectic alloy subjected to this "homogeneous cycle" was undercooled to a substantially lower extent, slight-

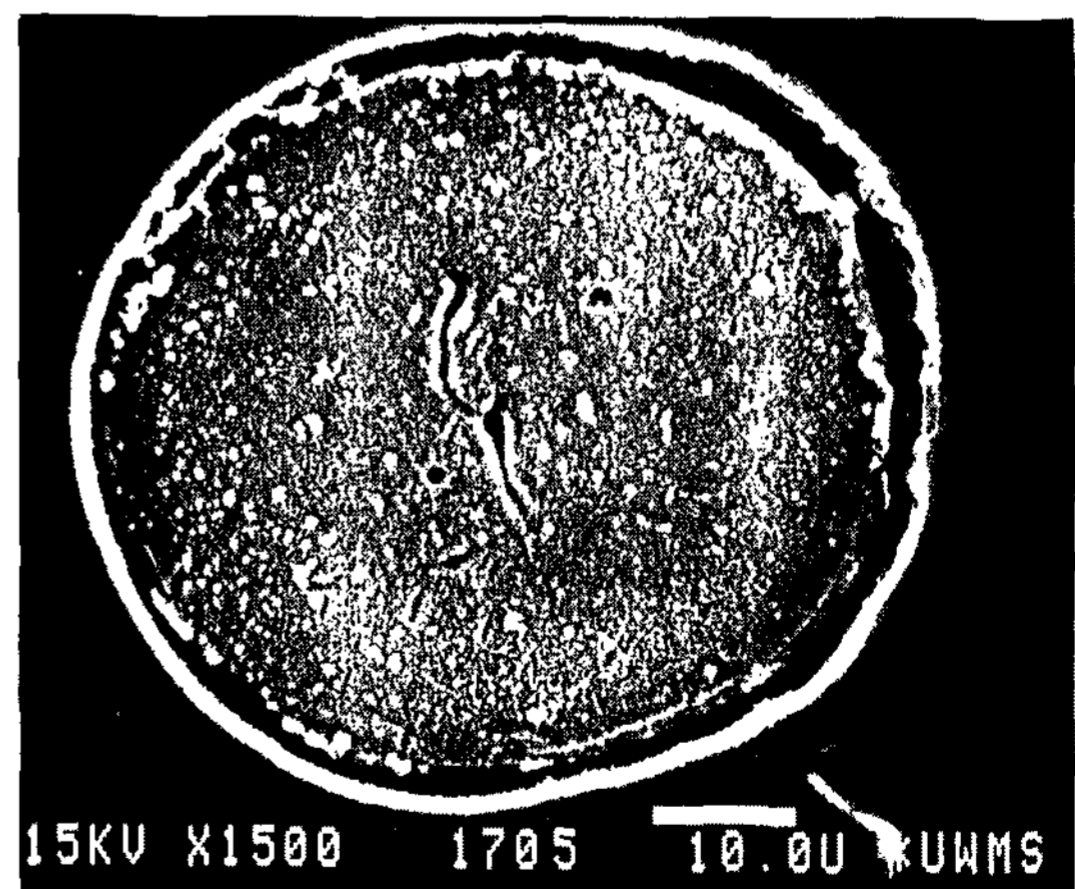


Fig. 12. Photomicrograph illustrating the microstructure of the 3.45% C, 1.91% Si emulsion which was supercooled after the "homogeneous" cycle to about 520°C and solidified as supersaturated austenite. Subsequent to solidification a fine dispersion of carbides precipitated throughout. This carbide precipitation following the solidification event was observed in other samples, with the volume fraction of carbide increased as the carbon content was increased.

ly more than 20°C.

The "heterogeneous cycle" was also applied to these alloys in order to assess the effectiveness of austenite and/or graphite as nucleants for the solidification of the eutectic. Under these conditions the 2.86% C alloy was undercooled significantly less than had been observed in the "homogeneous cycle", but to an extent that clearly demonstrates the fact that austenite was a poor nucleant of the eutectic. The 3.45% C sample exhibited similar results, Figure 13, with associated undercooling. The microstructure of the droplet consisted of the primary austenite phase plus a fine eutectic structure with lamellar spacing less than 1 μm and an average carbon content of the eutectic of about 9% (x-ray analysis). This is a new type of metastable carbide eutectic structure.

On the other hand the "heterogeneous cycle" applied to the hyper-eutectic alloy of 4.14% C (where the proeutectic graphite phase was present at the start of cooling) resulted in undercoolings of only about 20. Austenite formed a halo around the pre-existing graphite flakes except here the end of the flake was in contact with the liquid, Figure 14. At that location, graphite grew epitaxially from exposed basal planes of the primary graphite.

It was demonstrated in this study that primary austenite is a poor nucleant for both carbide and graphite, while graphite is an effective nucleant for austenite. This demonstrates quite effectively that interpretation of effective nucleants can not be based upon the lattice

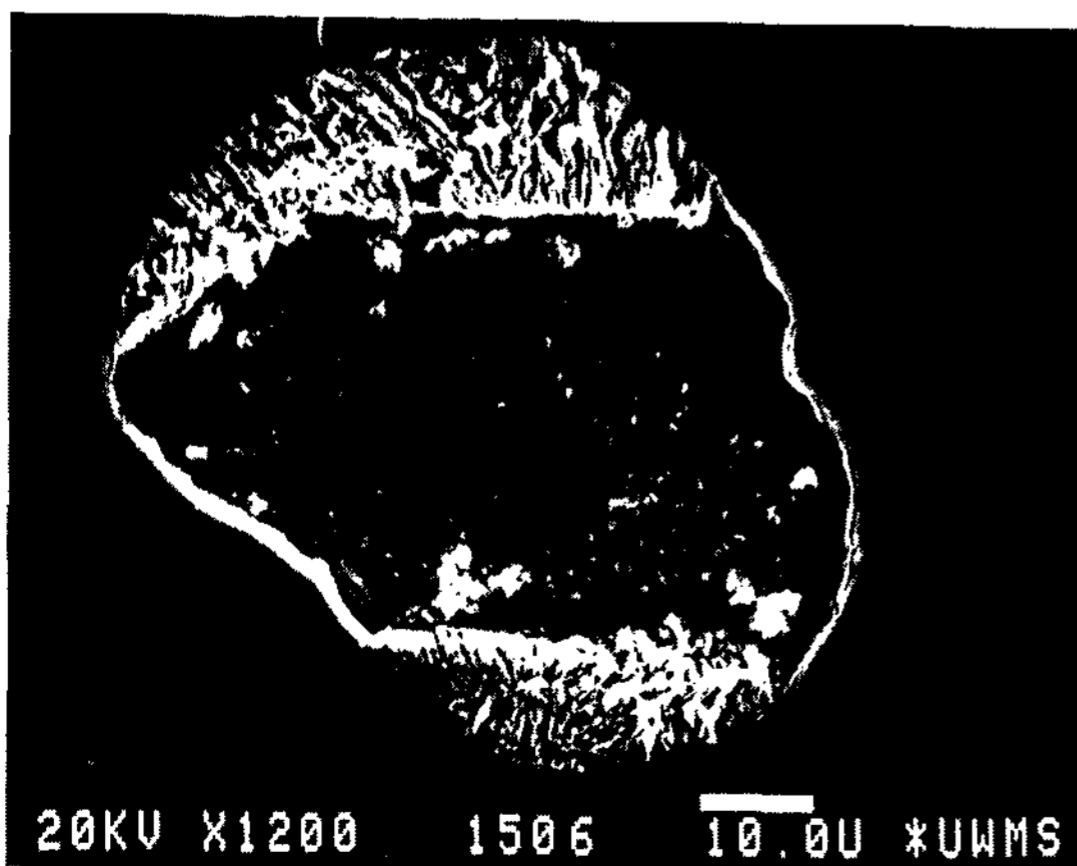


Fig. 13. Photomicrograph illustrating the microstructure of the 3.45%, 1.91% Si emulsion which was supercooled after the "heterogeneous" cycle.

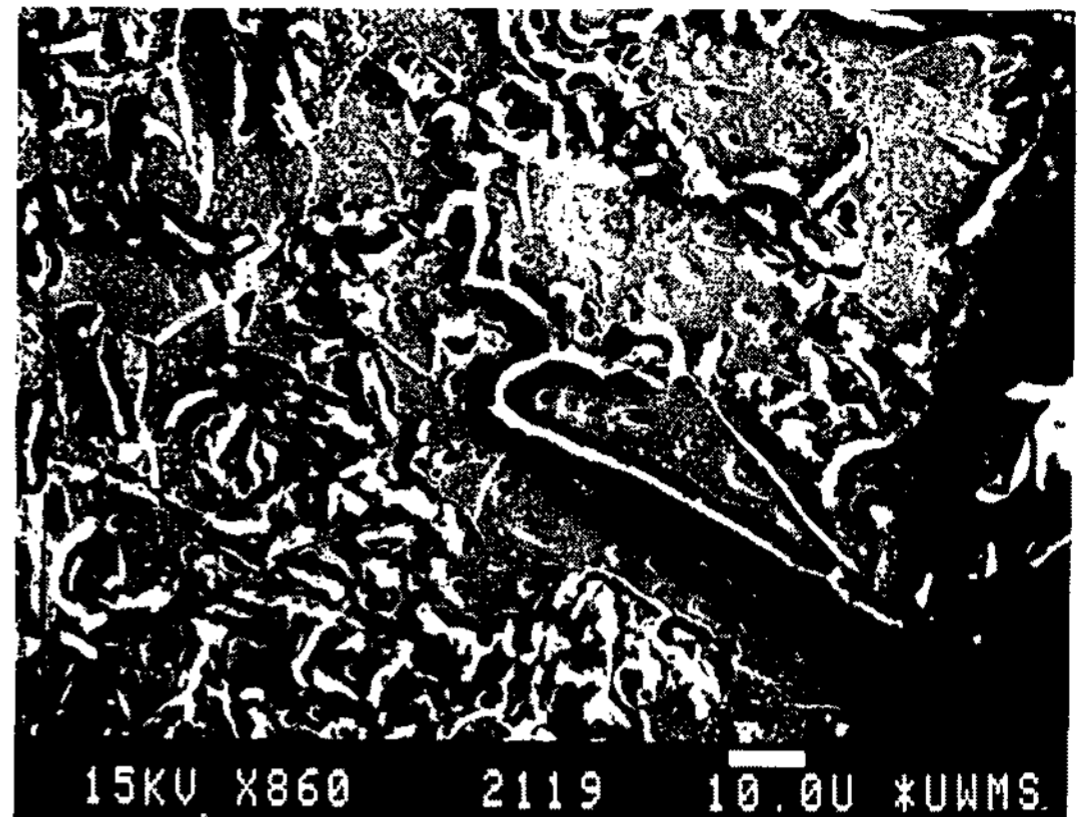


Fig. 14. Photomicrograph illustrating the microstructure of the 4.14% C, 1.91% Si emulsion subjected to the "heterogeneous" cycle. The large graphite flake was present in the droplet prior to cooling.

disregistry theory alone. Effective nucleation requires consideration of both the requirements of lattice disregistry between the nucleant and the nucleating solid and the relative interfacial energy between solids with respect to the melt.

Summary. Graphite has been demonstrated to be a highly effective nucleant for cast iron melts. However, its use must be undertaken only with an understanding of the manner whereby it is effective. Sufficient crystallinity of the graphite is essential in graphitic inoculants, as low crystallinity additives will not serve to nucleate the eutectic.

It follows, then, that suggestions made regarding an active role of austenite in the nucleation of the cast iron eutectic (stable or meta-stable) are not well founded. However, the formation of austenite from the liquid during slow cooling may result in development of carbon saturated liquid ahead of the solid-liquid interface and the resultant heterogeneous nucleation of graphite within that liquid. That graphite phase will then nucleate the graphite-austenite eutectic.

Fading of Inoculation Effectiveness

As noted previously, all inoculating effects will fade with time, the fading rate being more rapid at higher melt temperatures. Fading may be attributed to the agglomeration of inclusions in the melt, or to the dissolution of the inoculant addition in the melt. Examples of fading effects of various inoculants added to

gray cast irons and poured into standard ASTM chill molds are well known.

It is of significant interest to note that combinations of foundry grade FeSi and high purity graphite exhibited a much slower fade rate than was observed using either of these materials alone. The use of combined inoculating additions has been a well proven practice in specialty foundries where avoidance of chill is essential.

It may also be noted that graphite containing impurities, or "alloy" additions, exhibits markedly reduced fading characteristics. This fact should provide considerable opportunity for the development of speci-

alized graphitic inoculants which are effective chill reducers, but where their dissolution is controlled or inhibited by proper material selection or material production.

Acknowledgments

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國內外鑄物關聯行事

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(Modern Foundry
Technologies-Enviromental
Protection)
Cracow, 폴란드

9월 23일 ~ 9월 25일
Fonderie '97
Paris, 프랑스

9월 23일 ~ 9월 25일
제5차 아시아 주조대회
Nanjing, 중국

9월 30일 ~ 10월 1일
체코주조대회
Brno, 체코

10월 22일 ~ 10월 24일
제131차 일본주조공학회
전국강연대회
Hukushima, 일본

10월 21일 ~ 10월 24일
CASTING-97(Exhibition Fair)
Kiev, UKRAINE

11월 2일 ~ 11월 5일
ARABCAST-1997
Alexanderia, 이집트

11월 3일 ~ 11월 6일
제19차 국제 다이캐스팅
대회 및 전시회
Minneapolis, 미국

11월 6일 ~ 11월 7일
한국주조공학회 정기총회
추계 학술대회 및 기술강연 대회
교육문화회관, 경주

11월 26일 ~ 11월 29일
SHIN-SOKEIZAIFAIR '97
Yokohama, 일본