3.3 Austenite dendrites in SG iron

In 1961 Karsay first put forward the phenomena of austenite dendrites in heavy section spheroidal graphite iron castings[78]. However, for many years, the research work on austenite dendrites in SG iron was not as deep and extensive as that in grey iron. In recent years, with the technological developments of SG iron, the understanding of austenite dendrites has been improved and enhanced. Van de Velde has done the investigation on austenite dendrites for a long time [61, 79-80], he pointed out that previous spheroidisation theories only considered graphite itself, not the influence of austenite. He considered that the morphology of austenite is a driving force for the formation of other structures including graphite; the branching and growth of austenite influence the final morphology, size and distribution of graphite spheroids. Large dendrite spacing will cause large graphite spheroids, whilst small spacing leads to small spheroids, because part of graphite is precipitated in the liquid between dendrites. He also considered that the control of austenite quantity and morphology is an important step to obtain high quality SG iron. Although these viewpoints need further investigation and proof, it is certain that the effect of austenite dendrites should not be neglected in the research on structure, quality, and properties of SG iron. The research results of the author also confirm that austenite has an important influence on the solidification process and that austenite is an important solidification phase in the structure of SG iron.

3.3.1 Classification of austenite dendrites in SG cast iron

SG iron is a type of ‘divorced eutectic’; during eutectic crystallisation, austenite and graphite nucleate independently of each other and at different locations respectively. Therefore, except for primary austenite dendrites, divorced eutectic austenite will form different dendrite morphologies under different formation conditions. According to their growth conditions and morphology features, the classification of austenite dendrites is listed in Table 3-12.

3.3.2 Formation of austenite dendrite

(1) Primary austenite dendrite

For solidification of hypoeutectic SG iron, whether under equilibrium or non-equilibrium conditions, pre-eutectic or primary austenite will precipitate first, following the rules of nucleation and growth of austenite in Fe-C alloy, (see 2.3 in chapter 2).

(2) Chilled dendrite

On the surface of chilled SG iron castings, dendrites, which are perpendicular to the chilled surface, can form and these are called chilled dendrites.

Chilled dendrite belongs to pre-eutectic austenite and is a type

<table>
<thead>
<tr>
<th>Type</th>
<th>Growth condition</th>
</tr>
</thead>
<tbody>
<tr>
<td>Primary dendrite</td>
<td></td>
</tr>
<tr>
<td>Primary austenite dendrite</td>
<td>Hypoeutectic composition</td>
</tr>
<tr>
<td>Chilled dendrite</td>
<td>Fast cooling</td>
</tr>
<tr>
<td>Eutectic dendrite</td>
<td></td>
</tr>
<tr>
<td>Divorced dendrite</td>
<td>Eutectic structure</td>
</tr>
<tr>
<td>Halo dendrite</td>
<td>Round graphite spheroid</td>
</tr>
<tr>
<td>Slow cooling dendrite</td>
<td>Slow cooling, eutectic or hypereutectic</td>
</tr>
<tr>
<td>‘Spiking’ structure</td>
<td>Hypereutectic, small and thin casting, LTF</td>
</tr>
</tbody>
</table>
of primary dendrite. They can form in hypoeutectic and even hypereutectic SG iron. Under non-equilibrium freezing and fast cooling (see Fig. 1-13), hypereutectic SG iron will be undercooled to point 2 in the hypoeutectic region at left to eutectic region, causing austenite dendrites to precipitate. Because of fast cooling, heat will dissipate along the direction perpendicular to chilled wall, and the dendrites will grow following thermal flow direction. The primary dendrites are long and thin (can reach 50 mm), and have no branching, see Fig. 3-18.

(3) Divorced dendrite

When SG iron enters the eutectic solidification stage, eutectic austenite and graphite form at separate locations (see Fig. 3-19) and the austenite is known as ‘divorced dendrites’. Experiments have found that with increasing cooling rate, the formation of divorced dendrites increases; divorced dendrites have less apparent orientation and a more developed second axis than chilled dendrites.

(4) Halo dendrite

Halo dendrite is a type of divorced austenite forming around graphite spheroids; the halo or shell envelops spheroids with dendritic grown austenite, see Fig. 3-20. The formation process is described in section 3.4 of this chapter.

(5) Slow cooling dendrite

In thick section (>100 mm) SG iron castings of eutectic or hypereutectic composition, the dendrite morphology in Fig. 3-21 can be observed by the naked eyes; since the dendrite is produced under slow cooling conditions, it is named a ‘slow cooling’ dendrite, or ‘high temperature’ dendrite [81]. The main stem of slow cooling dendrites is perpendicular to the mould wall, bending upwards along the heat flow direction; the longest arm can reach 70–80 mm; the second dendrite arm is more developed than that in a chilled dendrite, but still arranged mainly as primary dendrite arm. In the centre of the section, the heat flow direction is not apparent, thus dendrites present randomly distributed equal-axis form. There exist graphite spheroids within slow cooling dendrites; the microstructure is shown in Fig. 3-22. The process of graphite precipitates along slow cooling dendrites is illustrated in Fig. 3-23.
Formation mechanism of slow cooling dendrite:

1. Undercooling theory: Mulazimoglu et al. proposed that the austenite dendrites formed in SG iron are related to the shortage of nuclei in the liquid iron. Due to various reasons (such as high superheating temperature, low residual Mg or Ce, or too low sulphur content), the liquid iron is short of nucleation substrates during eutectic solidification and the eutectic reaction is forced to occur at a lower temperature, causing much undercooling. The original hypereutectic SG iron is transformed to hypoeutectic, thus austenite dendrites are precipitated first.

2. Divorced eutectic theory: Loper considered that the austenite dendrite formed at high temperature is a necessary behavior of divorced eutectics and is a type of dendrite eutectic structure formed at close to or a slightly above the eutectic temperature. When the cooling rate is slow, (in thick section castings, for example), dendrites have enough time to grow and the dendritic structure can be clearly seen. Under the cooling conditions encountered in medium-section thickness castings, dendrites grows mainly in the halo-dendrite mode.

The author considered that in addition to the feature of divorced eutectic, the presence of slow-cooling dendrites is closely related to the nucleation status of the liquid iron. When the liquid iron has less graphite nuclei, but good nucleation conditions for austenite, the divorced austenite will precipitate first at high temperature. For heavy section SG iron castings, with slow cooling rate, solidification time is long and the number of graphite nuclei is significantly low, thus the dendrite precipitation tendency increases. For SG iron with high C and Si, even if the section is not very thick, it is possible to form obvious dendrites. This is because during crystallisation of hypereutectic SG iron, after flotation of primary graphite, the carbon content in the iron melt is reduced, which is favourable for austenite nucleation.

A slow cooling dendrite is a type of divorced eutectic, formed at an early stage of eutectic solidification; the latent heat of crystallisation released from austenite precipitation raises the eutectic temperature and causes a high temperature plateau at 1,170–1,190 °C on the cooling curve. After that, the iron melt enters the main eutectic reaction stage and forms a eutectic plateau. Therefore, the presence of two temperature plateaus is an indication of the formation of slow-cooling dendrites. The microstructure feature of slow cooling dendrites is that graphite spheroids are arranged according to the dendrite morphologies. This is because the carbon supersaturation in the regions close to divorced dendrites causes graphite to nucleate and grow around dendrites, then to be gradually enveloped by austenite, thus retaining graphite spheroids in a dendritic distribution.
In hot spot areas, (such as feeder, feeder neck and the connection between ingate and casting), of thick section (20–60 mm), hypereutectic SG iron castings, sometimes irregular black spot structure can be observed on the macro-fracture surface (see Fig. 3-24); this structure is also called spiking fracture\(^\text{(85)}\). Observation under a low magnification (x 30) microscope shows that the black spots consist of one or more large austenite dendrites, among which the spiking with a definite orientation are distributed. After etching, it is revealed that there exists regular ferrite bands, with graphite spheroids arranged in lines, see Fig.3-25. Because the austenite grain size is large and the direction of heat dissipation is not obvious, the primary and secondary axes of dendrites form long and straight crystal flakes, with graphite spheroids arranged along the dendrite interface to form spheroid lines (or chains).

Spiking is formed at the eutectic stage due to the formation of a large, divorced austenite dendrite that normally appears in a high temperature area, which is the last to freeze.

Similar to slow cooling dendrites, there exists a high temperature plateau on the cooling curve with the presence of spiking, see Fig.3-26. The reason why spiking is formed is related to cooling rate and nucleation state.

If in some area, there are less austenite nucleation substrates, austenite can only grow from the lower number of substrates; because of the high growth temperature, coarse and large austenite dendrites are formed, resulting in spiking structure. Although it is not clear why the number of austenite nuclei becomes less in certain areas of hot spots in thick section castings, the improvement of austenite nucleation conditions can effectively inhibit the formation of spiking. For example, Ti forms titanium carbide nitride Ti(CN), thus promoting austenite nucleation and fine austenite grains, which retard spiking; at the same time, a high content of sulphur, magnesium and rare earth metal can also reduce the probability of forming spiking. Besides, the formation of spiking is related to casting size and shape and by changing the location of the ingate and/or feeder ‘spiking’ may be increased, reduced, eliminated or moved to another location. The formation mechanism of spiking is more complicated than that of slow cooling dendrites.

3.3.3 Factors affecting the slow cooling austenite dendrite

Factors affecting on primary austenite dendrite in hypoeutectic SG iron are similar to that of grey iron\(^\text{(86)}\). With decreasing carbon equivalent and section thickness, the amount of austenite dendrites is increased. When increasing pouring temperature and silicon content, the amount of austenite dendrites is decreased.

However, for slow cooling dendrites in SG iron, the observed rule is obviously different. With increasing section thickness, slow cooling dendrite is more developed. In SG iron with wall section above 100 mm, the frequency of dendrite presence is obviously increased. With increasing carbon saturation degree, the critical section thickness decreases. In SG iron with a carbon equivalent of 4.6%, more slow cooling dendrites are likely to appear in areas with a section thickness of 60 mm (or above)\(^\text{(87)}\).

Although dendrites are possible to present both in eutectic and...
hypereutectic SG irons, the probability in hypereutectic SG irons is greater than in eutectic SG irons. For the same carbon content, silicon is a sensitive factor influencing the amount and morphology of dendrites, see Fig. 3-27. With increasing silicon, the shape of austenite dendrites becomes large and coarse, and the amount is increased; but when this leads to precipitation of chunky graphite, the amount of dendrites decreases again.

A high pouring temperature favours the formation of slow cooling dendrites.

Oxidising melting conditions, for example, a high quantity of moisture or a highly oxidised furnace charge, will increase the formation of dendrites.

The effect of trace elements on austenite dendrites is different; when Pb, Sb or Bi is present in iron melt, dendrites are significantly increased. When treating the iron melt with inoculants containing cerium, the nodule count increases, whilst the amount of dendrites decreases. By adding Sb to iron melt, the author did not eliminate austenite dendrites in heavy section SG iron castings; whilst adding B and Ti, the slow cooling dendrites are significantly decreased, see Fig. 3-28.

But, if the carbon saturation degree is too high, the effect of Ti is restricted. Besides, the addition of w(Bi) = 0.01% or Te can effectively inhibit dendrite formation. Effective inoculation can increase the number of graphite spheroids, promote formation of halo dendrites, and thus reduce the precipitation of slow cooling dendrites.

3. 3. 4 Effect of austenite dendrites on shrinkage porosity

Precipitation of austenite dendrites during the solidification process of SG iron exerts a marked effect on the type and formation mechanism of shrinkage porosity. The author considered that macro porosity often occurs in gaps between the dendrite clusters, whilst micro porosity often occurs between the dendrite arms or eutectic cells. The former occurs in the centre of hot spots, but the later can be distributed throughout all sections, although micro porosity is more prone in heavy section castings. It has been pointed out that macro porosity is formed at an early stage of eutectic solidification, after the formation of dendrite frames, and is the result of solidification contraction in other regions, which draw liquid from the hot spots. Micro porosity is formed at the final stage of solidification due to the solidification contraction of the last liquid between dendrites, which is not able to be fed.

3.3.5 Effect of austenite dendrites on mechanical properties of SG iron

The orientation arrangement of dendrites will cause a band-structure distribution, which will result in a difference in properties, when tested in different directions. Tests were carried out on a group of samples in which all the primary austenite dendrites were distributed with a particular orientation; some tests were carried out with the primary dendrite axis parallel to load direction and other tests were carried out with the primary dendrite axis perpendicular to load direction. Test results showed that when tension was applied parallel to the primary axis, mechanical properties of $\sigma_b = 570$ MPa, $\delta = 12\%$ were obtained, but when tension was applied perpendicular to the primary axis, mechanical properties were reduced: $\sigma_b = 464$ MPa, $\delta = 15.7\%$. This shows that the strength with parallel loading is 23% higher than that with perpendicular loading. The reason for the higher strength when tension is applied parallel to the dendrites is that for a static tension test, the loading speed and deformation are slow, thus deformation can spread throughout the section. When tension occurs along the primary dendrite axis, fracture is along the axis of the primary austenite dendrites; see Fig. 3-29(a). For impact tests, the results are different. When an impact load is applied parallel to the primary dendrite axis: $\alpha _k = 49.52$ J/cm$^2$ which is much lower than the value of 70.62 J/cm$^2$ obtained when the test is applied perpendicular to the primary dendrite axis. This is because under impact load, the load speed and deformation are rapid, and crack initiation and propagation occur easily in the regions between dendrites, which have poor plasticity; see Fig. 3-29(b).
If dendrites are distributed randomly, the mechanical properties will not show directionality. However, if dendrites can be arranged to favour the load direction, it is possible to develop the potential of these materials further.

Similarly, spiking decreases the mechanical properties of SG iron. Under load, the line-like arrangement of graphite spheroids and the presence of non-metallic inclusions produce stress concentrations and initiate cracks due to the notch effect. In addition, the ferrite bands caused by the line-like arrangement of graphite spheroids produce a weak region; thus a spiking dendrite structure decreases strength and elongation.

Fig. 3-28: The effect of B, Sb and Ti on slow cooling dendrites in SG iron

Fig. 3-29: Load direction and fracture location
3.4 Eutectic crystallisation of SG iron

Eutectic crystallisation of SG iron results in a ‘divorced eutectic’ and has different characteristics from that of grey iron. During eutectic solidification of SG iron, the relationship of graphite spheroids to austenite is different from that in grey iron. In grey iron, austenite and graphite have a close co-growth relationship, and graphite in the eutectic cell plays a prominent role. The freezing phenomena, which occur during eutectic crystallisation of SG iron, are:

1. Eutectic graphite spheroids and divorced austenite dendrites precipitate separately from each other at different locations.
2. Graphite precipitates around the austenite dendrites.
3. ‘Halo’ austenite nucleates and grows around the graphite spheroids.
4. The graphite spheroids grow within the austenite halo.

3.4.1 Divorced eutectic of SG iron

During eutectic crystallisation, the two phases of the eutectic, graphite and austenite, precipitate separately from the iron melt; their precipitation time is different and even some of the precipitation locations (but not all), are separated from each other; this is known as a divorced eutectic and is shown in Fig.3-30.

According to the principle of crystallization, a divorced eutectic easily occurs in the following circumstances:\n
(1) Non-equilibrium solidifying alloys with a strong hypo or strong hyper-eutectic composition

For alloys with a composition far away from the eutectic point, but still within the eutectic lines, the primary phase occupies most of the structure space; the eutectic liquid remaining in the gaps of the primary phase structures is rather small. Under non-equilibrium solidification conditions, one of the two eutectic phases grows, attached to the primary phase; the other phase precipitates rather later, in the spaces between the primary phase structures. Thus, the eutectic structure is not seen in the spaces between the primary phase structures, instead, only a certain single phase is present, thus forming a divorced eutectic.

(2) Sharp undercooling

Under severe undercooling conditions, one of the eutectic phases is difficult to nucleate and formation of one phase does not stimulate the formation of another phase, thus promoting the divorced eutectic. Severe undercooling due to the inherent properties of the alloy itself or fast cooling, will both favour the formation of a divorced eutectic.

(3) One of the eutectic phases enveloped by another phase

After forming an enveloping shell (called a halo), by a secondary phase, the growth of the primary phase is restricted within the halo; in order to continue the eutectic reaction, re-nucleation is necessary, thus resulting in a divorced eutectic.

After spheroidisation treatment with Mg and/or Ce, the undercooling of the iron melt is acutely increased and difficult to nucleate; because the surface of graphite spheroids is (0001) crystal plane, where the carbon atoms have the smallest atom spacing and are closely packed, it is very difficult for the spheroidal graphite to break through the austenite shell to grow, thus two phases with co-operative growth cannot form. Therefore, the SG iron conforms to the conditions of a divorced eutectic and solidifies in the divorced eutectic mode.

3.4.2 Formation of an austenite halo around a graphite spheroid

(1) Thermal and kinetic conditions of halo formation

When growing to a certain size, graphite spheroids are often enveloped by an austenite shell with a certain thickness; see Fig.3-31. The reason why it is easy to form a halo around a graphite spheroid is related to carbon concentration and also to the difference in growth velocity of the two phases. During graphite growth, the carbon distribution in the iron liquid varies; the carbon in the liquid around the graphite migrates towards the graphite to maintain graphite growth, thus forming a carbon-depleted region.

The carbon distribution in this depleted region is not uniform, the liquid closest to graphite has the lowest concentration; the liquid furthest from the graphite has a concentration near to the average of the melt, see Fig.3-32. The carbon-depleted region around the graphite offers a concentration condition for austenite nucleation; with the help of undercooling or heterogeneous nucleation, austenite nucleation is easily induced. The probability of nucleation is related to carbon concentration reduction (this refers to the carbon difference at the solid/liquid interface), while the carbon concentration reduction is mainly dependent on undercooling. Castings that are rapidly cooled show marked undercooling and an acute carbon concentration reduction, thus easily promoting austenite nucleation and accelerating the austenite halo formation. Heavy section castings show only a small degree of undercooling and small carbon concentration reduction around the graphite, thus it is difficult to form austenite halos.

In addition to carbon content, another important condition is the growth velocity difference between austenite and graphite. Only when austenite has a growth velocity greater than graphite, can it envelop graphite and form a halo. For ordinary grey iron, the growth velocity of a flake graphite tip is greater than that of austenite and austenite does not form a close halo. For liquid iron
(2) Formation process of austenite halo

Since there exist a good orientation relationship between the (0001) plane of graphite and the (111) plane of austenite, and a maximum carbon content reduction at the interface, it is easy to nucleate austenite on the surface of graphite, with graphite as the substrate \[ \text{[70]} \]. Soon after, austenite quickly grows in the region of reduced carbon content and forms a halo of certain thickness, that is, an austenite shell \[ \text{[92]} \]. Nevertheless, the formation mechanism of an austenite halo and the austenite growth model in the carbon-depleted region, have not been discussed in the literature. By using many experimental methods and observation of a large number of samples, the author proposed a description on the formation process, shown in Fig.3-33.

(a) Graphite spheroids nucleate and grow to a certain size in liquid iron.

(b) Carbon-depletion region forms around graphite, (the region is rich in silicon also).

(c) Austenite nucleates preferentially on the graphite (0001) plane which acts as substrate, see Fig.3-34. If heterogeneous nuclei exist outside the graphite interface, austenite can also nucleate there, see Fig.3-35.

(d) Austenite grows in a non-planar crystal mode, which should develop into dendrites; however, since the region around graphite has a high Si content, this makes the undercooling zone very narrow. In this case austenite is not in a suitable condition to branch laterally and form typical dendrites, but grows towards the liquid in a band-like, even blocky shape, see Fig.3-36. The released latent heat and the high-temperature environment cause band-like crystals to merge (or join each other), coarsen and form a whole austenite grain.

(e) Several austenite grains envelop a graphite spheroid; the austenite grains themselves then diffuse, homogenise and form an iso-compositional halo with annual growth-like rings.

The microstructure during the various formation stages of an austenite halo around a graphite spheroid is illustrated in Fig. 3-37.
3.4.3 Growth of graphite spheroids within an austenite halo

The further growth of a graphite spheroid after being enveloped by an austenite halo occurs by carbon diffusion through the halo. Since the diffusion velocity of carbon through this austenite shell is about twenty times less than that through liquid iron, the growth of graphite enveloped by an austenite halo is slow and the eutectic solidification time is prolonged. According to actual measurements, the growth time of spheroidal graphite directly in contact with liquid is very limited, thus the growth of spheroidal graphite is mainly completed within the austenite envelope. For an SG iron with carbon equivalent of 4.5%, the carbon precipitated (mass...
Fig. 3-36: Austenite (blue) grows in a band-like, or blocky shape within the halo.

Fig. 3-37: Microstructure during the formation process of an austenite halo.
fraction) before eutectic transformation is only 0.2%C (4.5%–
4.3%); whilst the carbon precipitated during eutectic transformation
is 2.0%–2.5%C. For normal heavy-section SG iron castings, the
graphite spheroids without an austenite shell are less than 10–15 μm
in diameter[35, 70], when just starting the eutectic growth, whilst at
the end of eutectic solidification, the maximum graphite diameter
can reach 50–100 μm. It can be seen that the growth of spheroidal
graphite during the eutectic stage is the dominant one.

The growth of graphite after being enveloped by an austenite shell
includes two key processes: (1) Carbon atoms diffuse from the
liquid towards a graphite crystal through the austenite shell. (2) Iron
atoms diffuse from the graphite/austenite interface, outwards.

After the graphite is enveloped by austenite, two interfaces, L-A
and A-G, form among the liquid, austenite and graphite. Figure
3-38(a) shows the carbon content at the interfaces. The carbon
content at the interface L-A is higher than that at A-G.

The carbon content at interface L-A is higher than that at interface
A-G, causing carbon to diffuse from interface L-A to interface A-G,
see Fig. 3-38(b). This oriented diffusion of carbon atoms within
the austenite shell, makes the carbon content at interface L-A lower
than that of the average content of the liquid iron, thus carbon in
the liquid phase will diffuse continuously into the austenite. Due
to oriented diffusion, the carbon content of austenite at the A-G
interface is increased and a carbon atom plane, which is parallel to
(111) plane of austenite, is gradually formed between (111) planes
of austenite. The atoms of the crystal plane are in equilibrium
and thermally vibrate constantly; when sufficient energy has been
obtained, the atoms will move to other positions. When approaching
freezing temperature, the transition becomes very frequent. As a
result, iron atoms constantly move away from the A-G interface
and leave vacancies for carbon atoms to crystallise to a new
(0001) plane. If spiral growth steps exist, this coherent growth will
accelerate.

In order to ensure that graphite grows continuously, vacancies
at the positions previously occupied by austenite, must be made
to let graphite “fill in”; thus the diffusion migration of iron atoms
is absolutely necessary. Because the diameter of a carbon atom is
much smaller than that of an iron atom, the diffusion velocity of
 carbon atoms is much faster than that of iron atoms. Therefore, the
diffusion of iron is a decisive factor influencing graphite growth.

### 3.4.4 Relationship between graphite shape and
austenite shell

When spheroidal graphite grows freely in liquid iron, the shape of
graphite is not influenced by the austenite shell, since austenite has
not yet formed. Once graphite spheroids are enveloped by austenite
shells, the final shape of graphite is inevitably restricted by the
austenite shells.

When precipitating as graphite within austenite shells, carbon
atoms will exert a greater expansion force on the austenite shells,
since graphite has a larger specific volume. When eutectic freezing
is nearing completion, eutectic cells contact each other and the
mould wall exerts a counter-pressure, increasing the effect of this
expansion force even further. The austenite shells deform under
this stress, resulting in high density of dislocation tangles and
deformation of the matrix structure, thus improving strength due
to an increased dislocation density. The higher the rigidity of the
mould, the better the strengthening effect [93].

It can be seen from Fig.3-33, that the austenite shells around
the graphite consist of several independent austenite grains; their
nucleation time and growth velocity are not the same, and this
influences the closure of the austenite shells. At an early stage,
austenite grains do not form a complete austenite shell; after
growing for some time, they contact each other and finally form
a closed shell. Before closure, liquid metal exists between the
austenite grains, which is called the liquid channel (also known as
ditch).

Enveloping can be divided into three types according to the
process involved [96, 97]:

1. **Fast enveloping:** When growing to certain sizes, austenite
   nucleates quickly in the carbon-depleted regions at the spheroidal
   graphite interface and quickly forms an enveloping austenite shell.
   Following this, the growth of spheroidal graphite depends on
   continuous carbon diffusion inwards. Because the austenite/graphite
   interfacial status is almost the same, the diffusion
   velocity of iron atoms is approximately the same all
   around. The homogeneous diffusion of carbon atoms
towards the surface of spheroidal graphite maintains
   the previous spheroidal growth. In most instances, this
type of shell is formed under fast cooling conditions.

2. **Slow enveloping:** When castings are slow-
   cooled, the iron, which is solidifying is short
   of sufficient nuclei and/or the liquid iron in the
   connecting channels is contaminated, the closure
   of austenite shells around the graphite spheroids is
   delayed; this leads to the formation of local gaps and
   results in a slow-enveloping shell.

3. **Non-enveloping:** When no nucleation in certain
   region around graphite spheroid, the region contacts
   with the last-to-freeze region constantly until final
   stage of freezing, this is called non-enveloping type.
   It is very difficult to give a strict definition for slow
and non-enveloping shells; it is also inconvenient to give an exact division between them.

The type of austenite shells is closely related to distortion or degeneration of graphite spheroids, see Table 3-13. Fast-enveloped austenite shells can maintain previously existing small, round graphite spheroids. A slow-enveloped austenite shell allows the graphite spheroid to be in contact with the liquid for a longer time, causing carbon atoms to diffuse easily towards the graphite spheroid; this leads to non-uniform growth of the graphite spheroid. For a non-enveloped austenite shell, the graphite spheroid is constantly in contact with liquid, through the liquid channel, until solidification is complete. Non-uniform diffusion of carbon towards a graphite spheroid and non-uniform restriction of austenite in the space, result in distortion of the graphite spheroid during the eutectic period.

The three types of austenite shell and their influence on graphite shape is shown in Figs. 3-39 to 3-41. In the solidified structure with low nodularity, there exist three types of shell, see Fig. 3-42. It can be seen clearly that all the shells showing good, round graphite nodules are the fast enveloping type, resulting in austenite shells, which are round and good. For poor shaped nodules, the austenite shells are not round or compact, and indicate the slow-enveloping or non-enveloping type.

The key factors influencing envelopment by austenite shells are:

### Table 3-13: Relationship between the type of austenite and graphite shape

<table>
<thead>
<tr>
<th>Austenite shell</th>
<th>Graphite formation process</th>
<th>Graphite shape</th>
</tr>
</thead>
<tbody>
<tr>
<td>(1) Fast enveloping</td>
<td><img src="image1.png" alt="Image" /></td>
<td><img src="image2.png" alt="Image" /></td>
</tr>
<tr>
<td>(2) Slow enveloping</td>
<td><img src="image3.png" alt="Image" /></td>
<td><img src="image4.png" alt="Image" /></td>
</tr>
<tr>
<td>(3) No enveloping</td>
<td><img src="image5.png" alt="Image" /></td>
<td><img src="image6.png" alt="Image" /></td>
</tr>
</tbody>
</table>

![Image](image7.png) **Fig.3-39: Fast enveloping austenite shells promote good graphite spheroids**

![Image](image8.png) **Fig.3-40: Effect of slow enveloping austenite shells on the distortion of graphite**
3.5 Crystallisation of the iron melt in LTF regions during the final stage of eutectic solidification of SG iron

During the final eutectic stage, the austenite shells, which envelop the graphite spheroids, gradually grow to a large size, the liquid iron between eutectic cells gradually becomes less and less, and the remaining liquid forms ‘LTF’ regions, as shown in Fig.3-43(a, b and c). The LTF regions in SG and grey iron are similar in chemical composition (high concentration of positive segregation elements) and freezing conditions; but, the features of LTF regions in SG iron are more evident and are more concentrated, and the formed structure has a more significant effect on mechanical properties compared with that of grey iron.
3.5.1 Intercellular carbides

Compared with grey iron, it is easier for carbides to form in SG iron \[\text{[96]}\]. The reasons are: (1) The ratio of surface area/volume of spheroidal graphite is at a minimum, thus decreasing the deposition velocity of carbon atoms onto graphite during graphite precipitation. (2) The spheroidising element Mg strongly promotes chill formation. (3) The liquid of SG iron has a low S content and high inoculation-fading tendency, therefore intercellular carbides (mainly Fe$_3$C) are easily formed in these LTF regions. Also, there exists a higher content of positive segregation elements Cr, Mo, V, Mn, and Ti in the LTF regions, which are easily combined with carbon and form various types of carbides, as shown in Fig. 3-44. If the iron melt has a high content of these elements, then large amount of carbides will form, resulting in a mottled structure; see Fig.3-45.

![Fig.3-43: Formation process of LTF regions in SG iron (dynamic observation of the same area)](image)

![Fig.3-44: Carbides in LTF regions](image)

3.5.2 Distorted graphite

When LTF regions contain an excessive amount of low melting point elements (such as Pb, As, Sn, Sb and Bi), intercellular flake (also called water sweet like graphite in China), even needle-like graphite, is easily formed \[\text{[97]}\], see Figs. 3-46(a) and 3-46(b). When the LTF regions contain excessive amounts of Ce, Ca or Sr, then chunky graphite is formed adjacent to these regions; see Fig.3-46(c).

3.5.3 Inclusions

Macro and micro inclusions often collect in the LTF regions and some of these are easy to identify using a standard optical microscope, (with colour etching method). For other inclusions, samples need to be electrochemically-etched (electrolyte: solution of glacial acetic acid containing CrO$_3$), and observed under a scanning electron microscope \[\text{[98]}\]. These inclusions are present in the form of compounds of Mg and Si or MgO, with a diameter of 1–3 μm, and are distributed in the LTF regions, at the intercellular interfaces. Shiao revealed that with an increase of w(Si) = 2.0% to w(Si) = 3.9%, the area of eutectic cells and the amount of inclusions both increase. It was also found that when under load, cracks propagate readily along inclusions, resulting in a decrease in toughness.
3.5.4 Austenite and its transformed structure

Due to the depletion of Si and the enrichment of Mn, Mo and Cr in the LTF regions, austenite in these regions becomes more stable; during eutectoid transformation, it is easy to form a pearlite structure in these regions, as shown in Fig. 3-47. With austempered irons, the austenite in LTF regions is very difficult to fully transform and some of the austenite is retained and/or transformed to martensite, resulting in ‘white areas’ in the structure; this reduces toughness, as illustrated in Fig. 3-48.

3.5.5 Shrinkage porosity

When liquid metal in the LTF regions freezes and cannot be fully fed, macro or micro shrinkage will form, see Fig. 3-49. Compared to grey iron of similar carbon equivalent, it is much easier for intercellular cavities to form in SG iron. This occurs because: (1) SG iron solidifies in ‘mush’ mode; the solidified shell is weak and cannot withstand the expansion which occurs during eutectic freezing; the casting expands outwards but does not compact inwards, thus can not form self-feeding. (2) The segregation tendency of carbide-promoting elements in the intercellular regions of SG iron is more pronounced than in grey iron.

The microstructures formed in LTF regions are closely related to chemical composition, cooling rate and nucleation status of the iron melt. It is not necessary that all the above five cases present. For most SG irons, stable austenite forms in LTF regions and transforms to pearlite. If it contains higher Si and lower Mn, this pearlite may further decompose and transform to ferrite.
Fig. 3-47: Pearlite structure in LTF regions: (a) and (b) are the same observation field

(a) Nital etch  (b) Hot alkaline etch

Fig. 3-48: Structure in ‘white area’ in an austempered SG iron

(a) Hot tint (240 °C, 5 h)  (b) Hot tint (300 °C, 5 h)

Fig. 3-49: Shrinkage porosity in LTF regions

(a) Micro porosity  (b) Macro porosity

(c) Macro porosity (under scanning electron microscope observation)

(To be continued)