Mechanism of the Peritectic Phase Transition in Fe–C and Fe–Ni Alloys under Conditions Close to Chemical and Thermal Equilibrium

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In-situ studies of the peritectic phase transition were performed under conditions close to chemical and thermal equilibrium in order to investigate the governing mechanism of both the peritectic reaction and the subsequent peritectic transformation using high-temperature laser-scanning microscopy in combination with the concentric solidification technique. Experiments have been conducted in the Fe–C and Fe–Ni systems in order to determine the role of solute diffusion on the observed reaction and transformation kinetics. The peritectic reaction as such as well as the subsequent peritectic transformation can be explained by diffusion-controlled mechanisms. Observations of the triple point L/γδ revealed that the peritectic reaction can be described as the solidification of γ along the L/δ interface, and new insights have been gained on the localized re-melting of δ during the peritectic reaction.

KEY WORDS: peritectic; solidification; phase transformation; steel.

1. Introduction

Over the past decades extensive research has been conducted to study the behavior and mechanism of the peritectic phase transition in different alloying systems. Because of a combination of fundamental scientific interest and technological importance, a large part of these investigations has been performed in Fe–C alloys and low- to medium alloyed steel types. In the Fe–C system, a peritectic is observed at 1768 K where δ-ferrite and liquid are in equilibrium with γ-austenite (further referred to as δ, L and γ). Here, peritectic solidification starts with primary precipitation of δ at some temperature below the liquidus temperature, which then solidifies until it reaches the peritectic temperature, TP. Under equilibrium conditions, γ nucleates just below TP and all three phases are in equilibrium, according to Gibb’s phase rule at constant temperature and pressure. Because the terms peritectic reaction, transformation and transition are still used inconsistently in the literature on peritectic systems, the following terminology, introduced by Kerr et al.1 will be used to distinguish between these different events. The peritectic phase transition initiates with the nucleation and growth of γ along the L/δ interface, this event being referred to as the peritectic reaction. The subsequent thickening of γ and the accompanying growth of δ into L and γ, respectively, are referred to as the peritectic transformation.

The development of new experimental techniques such as high-temperature laser-scanning confocal microscopy (HTLSCM) enabled, for the first time, in-situ observations and quantification of the peritectic reaction as distinct from the peritectic transformation.2–4 However, in earlier investigations experimental observations were all too often analyzed by over-simplified analytical and/or numerical models. The conclusions drawn from such approaches led to a variety of proposed mechanisms and diverse proposals for the governing mechanisms underpinning this industrially important phase transition. For example, the rate of the peritectic reaction in a Fe-0.14 wt.%C alloy was experimentally determined to be in the range of 6 mm/s for an undercooling of 5 K,2 which was much larger than the values calculated using Bosze and Trivedi’s simplified model of the peritectic reaction.5 Thus it was concluded that the peritectic reaction cannot be governed by a diffusion controlled mechanism. A new mechanism for the peritectic reaction was proposed by Phelan et al.,6 who suggested that the peritectic reaction is controlled by the rate of dissipation of heat of transformation that is released by the growing γ phase. In a different approach, Nassar et al.7 argued that the shape of the γ phase near the triple point L/γδ during the peritectic reaction is determined by the high surface tension at the γδ interface, which is a result of the elastic strain between γ and δ. They further argued that this strain causes an increase in Gibbs free energy, leading to an undercooling below the equilibrium peritectic temperature, thus explaining the high observed reaction velocities. In a recent study, Ohno et al.8 investigated the peritectic reaction in carbon steel by means of quantitative phase-field modeling. Their
calculated rates of the peritectic reaction were in fair agreement with the experimentally determined values available in the literature and thus, they concluded that the observed high reaction velocities can be explained by a carbon diffusion-controlled mechanism. In order to gain new insights into the governing mechanisms of the peritectic reaction as well as the subsequent peritectic transformation, experiments in the present study were designed to observe in-situ, the peritectic phase transition very close to equilibrium conditions.

2. Experimental Methods

Details of the high-temperature laser-scanning confocal microscopical technique (HTLSCM) used in this study have been described in detail in the literature and there is no need to repeat the details. Briefly, and pertinent to the present discussion, a He–Ne laser beam with a wavelength of 632.8 nm is scanned two-dimensionally (15.7 kHz × 60 Hz) and directed through a beam splitter and an objective lens before hitting the surface of the sample. The sample is placed in a gold plated, ellipsoidal shaped infra-red heating furnace under an ultra-high purity inert atmosphere (typically >99.9999% Ar). A 1.5 kW halogen lamp located at one focal point of the ellipsoidal cavity heats the specimen positioned at the other focal point. The temperature is measured by thermocouples incorporated in the crucible holder and simultaneously recorded with the image at a rate of 30 frames per second. Different sample geometries can be used, but the so-called concentric solidification technique has been demonstrated in an earlier study to be the most suitable arrangement for the study of the peritectic phase transition.

Figure 1 shows a schematic of the experimental setup, which is described as the formation of a centralized pool of liquid contained by a rim of solid of the same material under a radial thermal gradient. The temperature at the liquid/solid interface can be determined via temperature calibration.

The initial setup for the experimental investigations close to equilibrium conditions is schematically illustrated in Fig. 2. Specimen (9.8 mm diameter and 250 μm thickness) were heated to a temperature \( T \) (i.e. initial temperature) within the two-phase region \((L + \delta)\), in order to form a liquid melt pool with a radius, \( R(T) \). By doing so the initial fraction of solid \( f_s(T) \) before cooling is slightly less than the equilibrium fraction of solid \( f_s(T_P) \) at the peritectic temperature, \( T_P \), which corresponds to the pool radius \( R_E(T_P) \). The accuracy of the technique allows the establishment of a pool radius \( R \) which is only about 100 μm larger than the equilibrium peritectic pool radius, \( R_E \). Therefore, the established interface temperature is only about 1 to 2 K above the equilibrium peritectic temperature.

Following an isothermal holding period at the initial temperature to establish homogeneous conditions across the specimen, the temperature was slightly decreased manually, leading to a slow progression of the \( L/\delta \) interface until the equilibrium peritectic pool radius, \( R_E \), was reached and the peritectic transition was observed. By doing so, very small solute concentration gradients \((\frac{dC}{dx} < 5 \text{ m}^{-1})\) and undercoolings \((\Delta T << 1 \text{ K})\) were established prior to the occurrence of the peritectic transition and hence, the system remained close to equilibrium conditions.

3. In-Situ Observations

As the system was kept close to equilibrium conditions at all times during the experiment, the kinetics of the peritectic reaction were low enough to allow a detailed study of the governing mechanism.

3.1. Peritectic Reaction

The first event observed following cooling to below the equilibrium peritectic temperature was the nucleation and...
growth of \( \gamma \), i.e. the peritectic reaction. \( \gamma \) nucleates at \( \delta \)-
grain-boundaries in contact with the liquid melt, as earlier observed by Arai \textit{et al.} \textsuperscript{3} This nucleation process occurs
repeatedly at the same nucleation site when the temperature is cycled above and below the peritectic temperature several times in a row. Multiple nucleation events at different \( \delta \)-
grain-boundaries around the liquid melt pool were also observed. Once a stable \( \gamma \) nucleus is formed, the emerging \( \gamma \) phase grows laterally along the \( L/\delta \) interface in both directions at equal rates, as was also reported by Arai \textit{et al.} in an earlier investigation. \textsuperscript{3} Of great significance is that these observations have been made in the Fe–C as well as in the
Fe–Ni systems.

In \textbf{Fig. 3} a comparison is made of the shape of the \( \gamma \) plate-
let in a Fe-0.43C and Fe-4.2Ni alloy respectively for different reaction velocities. The thickness of the \( \gamma \) platelet decreases at higher reaction velocities in both alloying systems. In all cases the velocity, shape and the tip radius of the advancing \( \gamma \) platelet remained constant during the progress-
on along the \( L/\delta \) interface under given experimental conditions. An important observation, the implication of which will be discussed below, was that in the Fe–C system, the \( \gamma \) phase was located predominately in the \( \delta \)-phase during the peritectic reaction, whereas in the Fe–Ni system the growth of \( \gamma \) occurred mainly in the liquid phase. Also, the closer the system was kept to equilibrium conditions, \textit{i.e.} the lower the rate of the peritectic reaction, the higher was the amount of re-melted \( \delta \) in front of the triple point \( L/\gamma/\delta \). When the experimental conditions around the triple point were changed to strongly non-equilibrium conditions, no remelting of \( \delta \)-ferrite was observed and the implications of this observation will be discussed below.

The overall thickness as well as the tip radius of the advancing \( \gamma \) phase decreased with increasing reaction velocities along the \( L/\delta \) interface. In \textbf{Fig. 4} the tip radius of the \( \gamma \)-phase observed during the occurrence of the peritectic reaction is shown as a function of the velocity of the pro-
gressing \( \gamma \)-platelet.

\textbf{Figure 5} shows a comparison between the shape of the \( \gamma \)-phase during the peritectic reaction in a Fe-0.43 wt.%C

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**Fig. 3.** Triple point \( L/\gamma/\delta \) in a Fe-0.43C and Fe-4.2Ni alloy for different reaction velocities \( v_\gamma \) of the \( \gamma \) platelet: a) \( v_\gamma = 0 \) \( \mu \)m/s, b) \( v_\gamma = 5 \) \( \mu \)m/s, c) \( v_\gamma = 36 \) \( \mu \)m/s, d) \( v_\gamma = 125 \) \( \mu \)m/s, e) \( v_\gamma = 510 \) \( \mu \)m/s, f) \( v_\gamma = 185 \) \( \mu \)m/s, g) \( v_\gamma = 255 \) \( \mu \)m/s and h) \( v_\gamma = 1400 \) \( \mu \)m/s.

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**Fig. 4.** Tip radius of the \( \gamma \)-phase during the occurrence of the perite-
tic reaction in a Fe-0.43C alloy as a function of the reaction velocity (progression of the \( \gamma \)-platelet along the \( L/\delta \) interface).
alloy (Fig. 5(a)) and a Fe-4.2 wt.-%Ni alloy (Fig. 5(b)) with similar reaction velocities, $v_\gamma$. In the Fe–C system, the advancing $\gamma$-phase can be clearly divided into two different regions, i.e. a region of solidified $\gamma$ and a region where the $\delta$ has transformed into $\gamma$. The transformation of $\delta$ to $\gamma$ starts immediately behind the tip of the advancing $\gamma$ platelet (i.e. triple point $L/\gamma/\delta$). The transition between these regions can be identified as a groove in the region of the former $L/\delta$ interface. The surface ridge in $\delta$ in front of the $\gamma$ platelet (dashed line in Fig. 5(a)) is believed to be resulting from the local re-melting of $\delta$. The fact that this effect is only visible in the direct neighbourhood of the re-melted $\delta$ supports this conclusion. In the Fe–Ni system, only the solidified $\gamma$ is present without any detectable growth of $\gamma$ into $\delta$. Similar to the observations made in the Fe–C system, the interface between the solidified $\gamma$ and the $\delta$ can be detected as a groove between these two regions.

During the propagation of the $\gamma$-platelet along the $L/\delta$ interface, localized re-melting of $\delta$ has been observed in the Fe–C as well as the Fe–Ni system. For the purposes of further investigation, the $\gamma$-platelet has been stabilized at a fixed position at the $L/\delta$ interface, followed by small incremental growth regulated by a small drop in temperature, as shown in Fig. 6. The ability to stabilize the $\gamma$-platelet at the $L/\delta$ interface essentially requires a temperature gradient along this interface, which was experimentally achieved by a slight eccentricity of the melt pool. Figure 6 shows the sequence of events that occurs during the incremental growth of the peritectic $\gamma$-phase in a Fe-0.43C alloy. A schematic illustration of the pertaining temperature cycle is also provided below the frames. In order to minimize bulk effects, the specimen was machined to a thickness of 0.18 mm. When the $\gamma$ platelet was held at a stationary position at the $L/\gamma/\delta$ interface (Figs. 6(a), 6(c), 6(e) and 6(g)), no remelting of $\delta$ could be observed and the phases were in equilibrium at the triple point $L/\gamma/\delta$. The first event observed when the temperature was slightly decreased by $dT$ was remelting of $\delta$ while the $\gamma$ platelet remained in a stationary position (Figs. 6(b), 6(d) and 6(f)). After a certain length $dL$ of $\delta$ has been re-melted, the $\gamma$ phase subsequently continued...
to grow into the re-melted gap (Figs. 6(c), 6(e) and 6(g)). This incremental growth has been repeated three times as can be seen from the surface depressions which resulted from the stationary positions of the $\gamma$-platelet as shown in Fig. 6(g).

### 3.2. Peritectic Transformation

The growth of the $\gamma$-phase into the liquid and $\delta$-phase (i.e. the peritectic transformation), was observed to begin after the liquid phase and $\delta$ were locally separated by $\gamma$ during the peritectic reaction. Figure 7 shows a sequence of events that occur during the peritectic transformation in a Fe-0.43 wt.-%C alloy at different reaction velocities. The micrographs taken at 0 seconds were taken right in front of the advancing $\gamma$ platelet, i.e. the tip of the $\gamma$ platelet is right below the bottom of the micrographs. The same region has then been cropped out of the subsequent frames in order to ensure a measurement of the position of the transformation interfaces at the same location of the micrograph. The position of the initial L/$\gamma$ interface is indicated by a black arrow at the bottom of the micrographs. The same region has then been cropped out of the subsequent frames in order to ensure a measurement of the position of the transformation interfaces at the same location of the micrograph. The position of the initial L/$\gamma$ interface is indicated by a black arrow at the bottom of the micrographs and the positions of the respective transformation interfaces are indicated by dashed black lines. The position of the transformation interfaces are plotted as a function of the square root of time in Fig. 8 for different reaction velocities in the Fe–C system.

### 4. Discussion

#### 4.1. Nucleation of $\gamma$

The initiation of the peritectic phase transition by the nucleation of $\gamma$ takes place at $\delta$-grain-boundaries that are in contact with the liquid phase. This observation has been made in all of the alloys that have been investigated and has also been reported by other researchers. The nucleation of $\gamma$ has always been observed very close to the equilibrium
4.2. Peritectic Reaction

The shape and the thickness of the emerging γ phase during the peritectic reaction depend on the migration rate along the L/δ interface, as shown in Fig. 3. Nassar et al.\textsuperscript{13} argued that the shape and the velocity of the γ platelet are determined by the high surface tension at the γ/δ interface, which is a result of the elastic strain between γ and δ. This however could not be confirmed by Ohno et al.\textsuperscript{9} who investigated the effect of the interfacial energy on the rate of the peritectic reaction by using phase-field simulation. In the present study, the increased surface tension between δ and γ due to strain effects could also not be confirmed. Figure 9(a) shows the shape of the γ platelet in a Fe-0.43C alloy that forms when the temperature is held constant and the platelet remains stationary at a fixed position at the L/δ interface. In establishing mechanical equilibrium at the triple point, the surface forces will tend to impose an equilibrium morphology (minimum energy) in which the sum of the forces will be zero. \textsuperscript{14} This condition will be satisfied when

\[ \sigma_{\delta\gamma} = \sigma_{\gamma L} \cdot \cos(\theta_1) + \sigma_{\delta L} \cdot \cos(\theta_2) \] \hspace{1cm} (1)

and

\[ \sigma_{\gamma L} \cdot \sin(\theta_1) = \sigma_{\delta L} \cdot \sin(\theta_2) \] \hspace{1cm} (2)

Using Eq. (1) with the experimentally measured values of \( \theta_1 = 38°, \theta_2 = 51°, \sigma_\delta = 0.204 \text{ J/m}^2, \) \textsuperscript{14} the value for the γ/δ interfacial energy was calculated as \( \sigma_{\gamma \delta} = 0.592 \text{ J/m}^2, \) which is in good agreement with the values reported in the literature.\textsuperscript{15,16} The interfacial energy at the L/γ interface was calculated to be equal to \( \sigma_{\delta L} = 0.461 \text{ J/m}^2. \) For an advancing γ platelet (Fig. 9(b)) the surface tension vectors do not represent the true surface tensions as the shape of the γ tip is affected by the remelting of δ ahead of the γ tip. This situation is clearly shown in Fig. 6, where the surface tension vectors are changing while the γ platelet stays at a constant position during remelting of δ. Also, a measurement of the true interfacial tensions based on such experimental observations is difficult since the tension vectors need to be measured in the direct vicinity of the triple point which is evidently limited by the resolution of the recorded frames.

The rate of the peritectic reaction depends on the thermal conditions (\textit{i.e.} cooling rate and undercooling below the equilibrium peritectic temperature) as well as the diffusivity of the solute elements. Higher migration rates resulted in thinner γ platelets as well as a smaller tip radius (Fig. 4). This observation can be explained with the schematic shown in Fig. 10. The lateral growth of the γ platelet along the L/δ interface (\( v_{\gamma L} \)), separating the L from δ, occurs at a higher rate than the diffusion controlled thickening of this layer (\( v_{\gamma \delta} \)) and hence, the thickness of the γ platelets is decreasing with increasing reaction velocities.

From Fig. 5 it appears that, depending on the diffusivity of the solute elements, the γ platelet consists of a region of solidified as well as transformed γ. In the Fe–Ni system, only the solidified γ was observed due to the low diffusivity of the substitutional nickel atoms, which delayed the start of the peritectic transformation of δ to γ. In the Fe–C system, the peritectic transformation was observed to start immediately behind the triple point L/γ/δ due to the very high diffusivity of the interstitial carbon in γ. A comparison between these different alloying systems leads to the conclusion that the peritectic reaction can actually be described as the solidification of γ along the L/δ interface, whereas the peritectic transformation is the diffusion-controlled growth of γ into the δ and the liquid phase, respectively. Ohno et al.\textsuperscript{17} investigated the morphology of the triple point for a model alloy system by means of quantitative-phase-field simulation, showing a strong effect of the solute diffusivity on the morphology of the triple point and the shape of the γ platelet. They showed that a high solute diffusivity in the solid leads to a pronounced growth of γ into the δ, whereas a low solute diffusivity results in a growth of γ predominantly in the liquid phase. These simulations are in excellent agreement with the experimental observations of the present study. Ohno et al.\textsuperscript{17} also investigated the influence of the partition coefficient on the morphology of the phases near the triple point, which is also in excellent agreement with our experimental observations.

4.3. Remelting of δ

When the conditions around the triple-point L/γ/δ were close to equilibrium, remelting of δ has been observed in both alloying systems. In the literature two different explanations are suggested for the cause of the remelting of δ. In
the diffusion model proposed by Hillert,\(^\text{18}\) the enrichment of solute in front of the advancing \(\gamma\) tip leads to partial remelting of \(\delta\). In the heat transfer model proposed by Phelan \textit{et al.},\(^\text{6}\) remelting of \(\delta\) is explained by the dissipation of the latent heat of fusion released during the growth of the \(\gamma\). It follows from Fig. 6 that for a constant temperature at the triple point \(L/\gamma\delta\) (i.e. equilibrium peritectic temperature) the \(\gamma\) platelet remained at a stationary position at the \(L/\delta\) interface and no remelting of \(\delta\) could be observed. The first event observed when the temperature was slightly decreased is the remelting of \(\delta\) in front of the triple point, while the \(\gamma\) platelet remains at a stationary position. In the next step, the \(\gamma\) platelet solidifies into the region of re-melted \(\delta\) and remains at a stationary position when the temperature is held constant again. This cannot be explained by the heat transfer model, as the \(\gamma\) platelet remains at a stationary position during the remelting of \(\delta\). The latent heat of fusion is a result of the energy release during the change from an atomic short-range order to a long-range order (i.e. solidification), and therefore no latent heat can be released by the \(\gamma\) phase at this stage. Better agreement with the observed behavior is found by assuming a mechanism controlled by solute diffusion. When the temperature is held constant at the triple point \(L/\gamma\delta\), all three phases are in equilibrium with each other and no diffusion will occur. A decreasing temperature causes a difference in chemical potential of the elements, which in turn results in a driving force for the diffusion of the solute and the iron atoms. In the Fe–C as well as the Fe–Ni system, the partitioning process and the resulting remelting of \(\delta\) occurs at a much higher rate than the solidification of \(\gamma\) related to the smaller driving force for the diffusion of iron atoms.

### 4.4. Peritectic Transformation

In the Fe–C system, the growth of \(\gamma\) into \(L\) and \(\delta\), i.e. the peritectic transformation, was observed to start immediately after \(L\) and \(\delta\) were locally separated by the \(\gamma\) platelet during the peritectic reaction (Fig. 5). In the Fe–Ni system, the lower diffusivity of nickel in \(\gamma\) resulted in a clear separation of the peritectic transformation from the peritectic reaction. A higher rate of the peritectic reaction is accompanied by an increased rate of the peritectic transformation, reflecting the driving forces for diffusion at a certain undercooling below the equilibrium peritectic temperature (see Fig. 8). Furthermore it was found that the time dependence of the interface propagation during the peritectic transformation follows the relation \(\partial x = K \cdot t^{\frac{1}{2}}\), where \(K\) is a constant. This observation has also been made by other researchers\(^\text{2,19,20}\) and supports the conclusion of a diffusion controlled mechanism of the peritectic transformation.

Fredriksson and Nylen\(^\text{21}\) derived an expression for the growth rate of the transformation interfaces during the peritectic transformation:

\[
\frac{\partial d_\gamma}{\partial t} + \frac{\partial d_\delta}{\partial t} = \frac{D_\gamma}{d_\gamma} \left( \frac{x^{\gamma L} - x^{\gamma \delta}}{x^{\gamma L} - x^{\gamma \gamma}} \right) + \frac{D_\delta}{d_\delta} \left( \frac{x^{\delta \gamma} - x^{\delta \delta}}{x^{\delta \gamma} - x^{\delta \gamma}} \right)
\]

where \(d\) is the thickness of the \(\gamma\) layer, \(\partial d_\gamma/\partial t\) and \(\partial d_\delta/\partial t\) are the growth rates of the \(\gamma\) interfaces towards the \(\delta\) and liquid phase, respectively. \(D\) is the diffusion coefficient of solute atoms in \(\gamma\). It should be noted that Eq. (3) is derived under the assumption of small concentration gradients in the \(\delta\) and the liquid phases, i.e. the equation is only valid close to equilibrium conditions. Figure 11 shows a comparison of the observed interface migration rates with the values calculated by Eq. (3). The corresponding frames are shown in Fig. 7. In the calculation, the initial thickness of the \(\gamma\) layer at the beginning of the peritectic transformation was chosen to be 0.1 \(\mu m\) and the diffusion coefficient of carbon in \(\gamma\), \(D_\gamma\), was set to \(8 \times 10^{-10} m^2/s\).\(^\text{22}\) The cooling rate was set to 0.5 K/min and the corresponding undercooling was approximated from Shibata \textit{et al.},\(^\text{2}\) who measured the relationship between undercooling and reaction velocity in a Fe-0.42C

![Fig. 11. Comparison between calculated and experimentally observed interface migrations rates in a Fe-0.43C alloy for two different reaction velocities \(v_\gamma\).](image)

![Fig. 12. Comparison between calculated and experimentally observed interface migrations rates in a Fe-4.2Ni alloy for a reaction velocities of \(v_\gamma = 480 \mu m/s\).](image)
alloy. The undercooling for the observed reaction velocities \( v_\gamma \) of 63 \( \mu \text{m/s} \) and 125 \( \mu \text{m/s} \) were set to 0.25 K and 1 K, respectively.

In case of the \( \gamma/\delta \) interface, the calculation is in good agreement with the experimental observations, indicating diffusion-controlled growth. However, the calculated values of the \( L/\gamma \) interface are lower than experimentally observed for both experiments. This can be explained by the presence of convection in the liquid phase, as Eq. (3) does not take into account the solute diffusion in the liquid phase. Yin and Emi\(^{23}\) studied the flow of a steel melt in a similar experimental arrangement and observed strong Marangoni convection in an Fe–C alloy during solidification using a cylindrical specimen and HTLSCM. A higher solute diffusivity in the liquid phase increases the transport of solute to and from the \( L/\gamma \) interface, therefore leading to increased migration velocities of this interface. The same effect was found in the Fe-4.2Ni alloy (Fig. 12), which is also in agreement with the observations made by Arai et al.\(^{3}\)

5. Conclusions

High-temperature laser-scanning confocal microscopy in combination with a concentric solidification technique was utilized to study the governing mechanism of the peritectic phase transition under conditions close to chemical and thermal equilibrium. Both the peritectic reaction and the peritectic transformation are governed by a diffusion controlled mechanism and strongly depend on the solute diffusivity in the solid phases. The peritectic reaction can be described as the solidification of \( \gamma \) along the \( L/\delta \) interface and can be clearly distinguished from the subsequent peritectic transformation. The start of the peritectic transformation of \( \delta \) to \( \gamma \) following the peritectic reaction is strongly influenced by the solute diffusivity in the \( \gamma \) phase. Remelting of \( \delta \) ahead of the advancing \( \gamma \) tip can also be explained by a diffusion controlled mechanism.

REFERENCES