Thermal analysis and solidification pathways of Mg–Al–Ca system alloys

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Abstract

Mg–Al–Ca alloys are creep resistant magnesium alloys with high application potentials. The solidification pathways and microstructure formation in this alloy system are still under discussion. In this paper, the solidification behavior of AZ91 and AM50 with Ca addition (AZC91x and AMC50x alloys) was investigated by a computer-aided cooling curve analysis (CA-CCA) system. Microstructure and phase identification were carried out by SEM and EDX analysis. The results show that the Ca-containing phase formation mainly depends on Ca content and Ca/Al ratio. With increasing the Ca/Al ratio these phases transform from Al2Ca to (Mg, Al)2Ca and Mg2Ca. Moreover, Ca addition decreases the liquidus temperature of Mg–Al alloys, but influences the solidus temperature in a more complex way. Increasing the Ca content also decreases the solid fraction at which dendrite coherency occurs. The relationship between solidification interval, dendrite coherency point, formation of Ca-containing phases and hot tearing is also discussed.

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Keywords: Mg–Al–Ca alloys; Microstructural formation; Solidification pathway; Thermal analysis

1. Introduction

Owing to the increasing demand of cast magnesium components for automotive applications, a wide variety of magnesium alloys and product forms have been developed [1]. AZ and AM series magnesium alloys, which offer good combination of room temperature strength and ductility, good salt spray corrosion resistance and excellent die castability, have been widely used in instrument panels, steering wheels and valve covers. However, all applications of these alloys are limited to components operating at temperatures lower than 400 K because of their rapid strength reduction at higher temperatures. In order to overcome these shortcomings, Ca-containing magnesium alloys have drawn many researchers’ attention in the last decade since Ca addition was proved to be not only beneficial to mechanical properties at elevated temperature but also Ca is cheap and light as alloying element [2,3]. The excellent creep resistance and elevated temperature properties of the Mg–Al–Ca alloys have been attributed to the formation of high melting temperature Ca-containing phases with various sizes located both at the grain boundaries and within grains interior [4,5].

Although high pressure die casting (HPDC) is by far the most popular process to manufacture magnesium components, sand casting has been used for a very long time for large aerospace castings such as helicopter gear box housings. However, it is well recognized that, whatever process is taken, the mechanical properties of the Ca-containing magnesium alloy components could not be improved further by heat treatment, because the size and the morphology of the secondary phases in these alloys remain almost unmodified under any heat treatment conditions. In addition, heat treatment is hardly carried out on die casting parts since air bubbles entrapped beneath the surface during the process would bulge and damage the surface of the components. Therefore, the properties of the components produced from these alloys would depend to a large extent on the selected alloy composition and on the microstructure resulting directly from solidification. However, the currently available Mg–Al–Ca

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ternary phase diagrams are still incomplete and inconsistencies exist so that any prediction of the equilibrium or non-equilibrium solidification structures for these alloys is impossible [6–8]. As a result, the understanding of the solidification behavior and microstructure formation pathways becomes a crucial issue to widen the applications of creep resistance magnesium alloys containing Ca addition.

Thermal analysis techniques, such as differential thermal analysis (DTA), differential scanning calorimetry (DSC), computer-aided cooling curve analysis (CA-CCA), are powerful tools to investigate the solidification characteristics of alloys, namely characteristic temperatures, solid fraction and microstructural evolution [9]. Computer-aided cooling curve analysis (CA-CCA), which is simple, inexpensive and most importantly suitable for industrial applications, has been applied efficiently for many years to investigate the development of microstructure during solidification of Fe and Al alloys [9–13]. The present research is conducted to understand in more details the solidification pathways of Mg–Al–Ca alloys (mainly based on AZ91 and AM50 with Ca addition) by a CA-CCA system. The liquidus and solidus temperatures, the variation of solid fraction with temperature, the second phase formation, and dendrite coherency point are then characterized. This characterization will lead to a better understanding concerning the formation of cast microstructures and defects and will help to widen the application of newly developed Ca-containing magnesium alloys.

2. Experimental procedures

Commercial high purity Mg ingots were melted in a 15 kg electrical resistance furnace using a mild steel crucible, then pure Al, pure Zn, Al–8 wt%Mn and Mg–30 wt%Ca master alloys are successively added into the melt at 700 °C; after stirring for 5 min, the melt was held for 10 min; after that, it was poured at 720 °C into a preheated steel mould with a diameter of 100 mm. Alloying elements contents were analyzed by using the ICP-AES method and the results of the analyses are presented in Table 1. Two AZ91 + Ca alloys (AZC911 with 1%Ca and AZC912 with 2%Ca) together with three AM50 + Ca alloys (namely AMC501 with 1%Ca, AMC502 with 2%Ca and AMC503 with 3%Ca) were prepared.

Samples for thermal analysis were cut from the same position of the cast ingots and machined into cylinders of 25 mm in diameter and 50 mm in length. The samples were remelted in a steel crucible in an electrical resistance furnace, then two shielded K-type thermocouples were immersed from the top of the crucible that was insulated from the top and the bottom. One thermocouple was placed at the center of the crucible (Tc) and the other close to the inner wall (Tn) at exactly the same depth in the melt (25 mm from the bottom). After holding 10 min at 720 °C, the crucible together with the two thermocouples were removed from the furnace and allowed cooling in air with a cooling rate of about 1 K/sec. The temperature changes were continuously recorded during the solidification process by using a high-speed data acquisition system linked to the computer. The thermal analysis was conducted three times for each alloy composition to ensure the reliability of the results; three thermal analysis results showed that the temperature deviations at critical points were always approximately 1 °C, indicating that the reliability of the thermal analysis results in the present study is very good. Samples for microstructural analysis were taken from the regions near the thermocouple tip after complete solidification. Observations of the microstructure were conducted on a Philips XL30 ESEM-FEG/EDAX scanning electron microscope (SEM). The compositions of the Ca-containing phases were analyzed by EDAX, and the reported values are the average of 10 measurements for each sample.

3. Results

3.1. Phase formation

Fig. 1 shows the thermal analysis results for the AZC91x alloys (x = 0, 1, 2) and their microstructures in the as-cast state. The first derivative of the cooling curve (dTc/dt) was determined to enhance slope changes that are related to the solidification reactions for the different phases, and to facilitate the determination of the critical solidification characteristics of the alloys. In the AZ91 alloy, two well-defined peaks are observed at 605 and 424 °C, which correspond to the primary Mg phase formation reaction and the non-equilibrium eutectic reaction, respectively, as shown in Fig. 1(a). In accordance with Fig. 1(a), the microstructure of the AZ91 alloy is composed of the α-Mg matrix and the partially divorced eutectic β-Mg17Al12 phase as shown in Fig. 1(b). When Ca is added, a new peak prior to the eutectic reaction occurs at 483 °C for the AZC911 and 514 °C for the AZC912 alloy, as shown in Fig. 1(c) and (e), which correspond to a new phase formation reaction taking place during solidification. The microstructures of AZC911 and AZC912 are shown in Fig. 1(d) and (f), respectively. With increasing Ca content in AZ91, the amount of the Mg17Al12 phase decreases and becomes completely divorced, whereas the amount of this new phase increases. The EDAX analysis of the region corresponding to this new phase gives Mg10Al60Ca30 (at%), which is in agreement with the composition of the Al2Ca phase [8].

Fig. 2 shows the measured cooling curves for the AMC50x alloys (x = 0, 1, 2, 3) and their microstructures in the as-cast state. Similar to AZ91, solidification of AM50, as shown in Fig. 2(a), leads to the formation of the primary Mg phase and to the non-equilibrium eutectic reaction, which occur at 623 and 433 °C, respectively. The microstructural observation shown in Fig. 2(b) agrees with the cooling curve analysis; the microstructure con-

Table 1

<table>
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<tr>
<th>Alloys</th>
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<th>Ca</th>
<th>Mn</th>
<th>Ca/Al</th>
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<td>1.03</td>
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<td>0.11</td>
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<tr>
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<td>1.02</td>
<td>2.05</td>
<td>0.32</td>
<td>0.23</td>
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<td>AM50</td>
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<td>–</td>
<td>0.30</td>
<td>0</td>
</tr>
<tr>
<td>AMC501</td>
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<td>–</td>
<td>0.62</td>
<td>0.28</td>
<td>0.12</td>
</tr>
<tr>
<td>AMC502</td>
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<td>–</td>
<td>1.63</td>
<td>0.32</td>
<td>0.32</td>
</tr>
<tr>
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<td>–</td>
<td>2.89</td>
<td>0.30</td>
<td>0.54</td>
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</tbody>
</table>
sists of α-Mg matrix and β-Mg_{17}Al_{12} distributed along grain boundaries. In AMC501, the cooling curve again shows two peaks at 501 and 421 °C after the first peak corresponding to the primary Mg phase formation reaction. From the microstructural observation shown in Fig. 2(d), it is confirmed that these two peaks can be attributed to the formation of the Al_{2}Ca and β-Mg_{17}Al_{12} phases. In AMC502, where the Ca level is 1.63 wt%, the cooling curve and first derivative curve shown in Fig. 2(e) markedly differ from those of the AM50 and AMC501 alloy; only two peaks can be seen at 614 and 526 °C. Microstructural observation (in Fig. 2(f)) suggests that the β-Mg_{17}Al_{12} phase has disappeared and has been replaced by a eutectic mixture composed of Mg and a new Ca-containing phase. The EDAX analysis shows that the composition of this new eutectic Ca-containing phase is Mg_{40}Al_{36}Ca_{24} (at%), which is similar to that reported by Suzuki et al. [6] and Amerion et al. [14], who designated this phase as (Mg, Al)_{2}Ca. As the Ca level further increases to 2.89 wt% (AMC503), beside the eutectic reaction involving the (Mg, Al)_{2}Ca phase, a new reaction is detected at 511 °C, as shown in Fig. 2(g), which corresponds to the Mg_{2}Ca formation as reported by Suzuki et al. [6]. The microstructure of AMC503 is shown in Fig. 2(h), the EDAX analysis of this phase
Fig. 2. Thermal analysis results and microstructures of AMC50x alloy: (a) and (b): AM50; (c) and (d): AMC501; (e) and (f): AMC502; (g) and (h): AMC503.
gives a composition of Mg$_{61}$Al$_{16}$Ca$_{23}$ (at%), which is within the composition range of Mg$_2$Ca phase as reported by Gröbner et al. [8] and Amerion et al. [14].

3.2. Solid fraction curves

The knowledge of solid fraction at different stages during solidification is critical to the success of some casting processes, e.g. semi-solid casting [9]. The cooling curves obtained from the thermocouple located in the center of the mold ($T_c$) were used to determine the critical reaction temperatures and solid fraction. The simplest and most widely used method of CA-CCA is the Newton thermal analysis method (NTA) [11–13], which is based on the assumption that there are no thermal gradients inside the sample during the cooling process. These conditions are most closely satisfied when a thin-walled metal cup is used and the cup is well insulated from both the top and the bottom during the cooling process as in the case of the present experimental procedure. Based on the commonly employed assumption in DTA and DSC analyses and on a similar work in commercial magnesium alloys conducted by Mirkovic et al. [15], the present study assumes that the volumic specific heat of alloy is independent of temperature and that the rate of release of latent heat per unit mass is the same for each phase formation occurring during the solidification process. Then, according to the heat balance during the solidification process, the solid fraction can be generated as follows:

$$f_s = \frac{\int_t^{t_s}[(dT/dr)_{cc} - (dT/dr)_{bl}]dr}{\int_t^{t_s}[(dT/dr)_{cc} - (dT/dr)_{bl}]dr}$$

where $cc$ refers to the cooling curve and $bl$ refers to the base line, $t_1$ and $t_s$ represent the time of the beginning and of the end of solidification, respectively, i.e. the time at the liquidus and the solidus.

The base line is the time evolution of the cooling rate of the alloy assuming that no phase transformation takes place. There are mainly two methods to determine the baseline: Newtonian analysis [11–13] and Fourier analysis [9,10]. It is reported that Fourier analysis prediction of the latent heat is more accurate than the Newtonian method, but the fraction of solid calculated by Newtonian and Fourier methods shows little differences in Al alloys [9]. The Fourier method, which should measure positions of two thermocouples at the beginning and the end of solidification, is much more complicated than the Newtonian method. Moreover, as thermal contraction of the alloys occurs during solidification, the exact positions of the two-thermocouple tips are difficult to measure. In the present work, The Newtonian base line has been estimated by curve fitting the base line data below the solidus and above the liquidus using a third order polynomial as developed by Kierkus and Solokowski [13]. The third order polynomial yields a correlation coefficient greater than 0.998. An example of polynomial is given below:

$$(dT/dr)_{bl} = A + BT + CT^2 + DT^3$$

where $A = -1.19$, $B = 1.49 \times 10^{-3}$, $C = 1.65 \times 10^{-7}$, $D = -2.03 \times 10^{-10}$ for AZ91, and $A = -1.40$, $B = 3.23 \times 10^{-4}$, $C = 1.32 \times 10^{-5}$, $D = -2.13 \times 10^{-8}$ for AM50. The calculated solid fraction curves for AZC and AMC alloys are shown in Fig. 3(a) and (b), respectively.

3.3. Dendrite coherency point (DCP)

The dendrite coherency point (DCP) defined for both a certain fraction solid $f_{DCP}$ and a certain temperature $T_{DCP}$ corresponds to the point of impingement of the dendrites where strength of the mush begins to develop [16]. The formation of microstructure and casting defects in die castings and permanent mold castings, such as dendrite size and morphology, hot tearing and porosity is suggested to be strongly dependent on dendrite coherency of the alloy which in turn depends on alloy composition and solidification conditions [17]. Therefore, the knowledge of DCP is essential to eliminate hot tearing and to understand microstructure formation during semi-solid forming and conventional casting processes [9,16].

Measurements of the dendrite coherency point have been carried out by different techniques over the past decades. Mechanical, i.e. rheological techniques and thermal analysis (TA) are the two main approaches to detect DCP. Rheological methods involve the rotation of a paddle or vane within a solidifying melt. The dendrite coherency point corresponds to the point where the strength increases gradually, owing to the progressive locking of the paddle by the developing interconnected dendrite network [9,18,19]. Thermal analysis is based on the assumption that the coherency point corresponds to a discontin-
ous increase of the thermal conductivity of the mushy material. The establishment of a dendritic network at the DCP will result in a rapid decrease in the temperature difference between the wall and the central regions, due to the higher thermal conductivity of the solid material compared with that of the liquid [19,21]. The thermal analysis method suggests that DCP occurs when the temperature difference between the wall (T_w) and center (T_c) thermocouples is at its first maximum during primary α-Mg solidification.

The results of the dendrite coherency solid fraction (f_{DCP}) determined by these two methods are widely reported [18–21]. Veldman et al. [19] showed that thermal analysis and rheological coherency results do not coincide. They argued that rheological determination is more accurate than thermal analysis since it is based on a direct measurement of a physical parameter. But these authors did not illustrate how they determined the solid fraction. On the other hand, Mackay et al. [21] suggested that thermal analysis might be more reliable as some speculations in rheological methods may influence DCP measurement, since rotating paddles can collide with dendrite tips before they contact with neighboring grains. These collisions may force mechanical locking of non-neighboring grains, therefore postponing coherency in the center of the cup. In addition, the temperature recording used during paddle experiment is located near the wall of the cup, while coherency is obviously occurring in the center of the cup. Our recent work suggested that the results obtained from these two techniques are comparable with each other in magnesium alloys [20] as observed also in aluminum alloys [18]. The above considerations therefore suggest that, thermal analysis is an acceptable way to determine the dendrite coherency point. In this work, the temperature of DCP (T_{DCP}) was determined by the two-thermocouple thermal analysis method as illustrated in Fig. 4. By examining both Figs. 3 and 4, the solid fraction at DCP for each alloy could be determined, as each temperature for DCP obtained from Fig. 4 corresponds to a solid fraction in Fig. 3. The variation of the solid fraction corresponding to dendrite coherency (f_{DCP}) as a function of Ca concentration is plotted in Fig. 5. The f_{DCP} value varies from 18 to 50 pct, and it decreases with increasing Ca addition in both series of alloys. The same trends are also reported in some binary aluminum alloys [18].

Fig. 4. Principle of the determination of the dendrite coherency point (DCP) by thermal analysis.

Fig. 5. The effect of Ca concentration on solid fraction at dendrite coherency.

4. Discussion

The Ca-containing phases existing in Mg–Al–Ca alloys are still under discussion. It was believed that Al_{2}Ca and Mg_{2}Ca are the main Ca-containing phases in Mg–Al–Ca alloys. Ninomiya et al. [22] examined Mg–(3–9) wt%Al–(1–5) wt%Ca alloys by X-ray diffraction and reported that, in the as-cast state or at temperatures below 400 °C, both Mg_{2}Ca and Al_{2}Ca phases exist in alloys where the Ca/Al mass ratio is greater than 0.8, whereas for smaller Ca/Al mass ratio, only the Al_{2}Ca phase is present. Tkachenko et al. [7] and Gröbner et al. [8] mentioned that a quasibinary eutectic L=α-Mg+Al_{2}Ca should occur at 535 °C in the Mg–Al–Ca quasibinary section with a eutectic point close to 79 at% Mg, whereas a ternary eutectic L=α-Mg+Al_{2}Ca+Mg_{2}Ca at approximately 9 at% Al and 79 at% Mg occurs at 510 °C. Recently, Luo et al. [4] found the existence of the (Mg, Al)_{2}Ca phase with a hexagonal structure in AMC503 alloy. Suzuki et al. [6,23] further investigated the (Mg, Al)_{2}Ca phase and identified this phase as a new Laves phase with a C36 structure that is different from Mg_{2}Ca (C14) and Al_{2}Ca (C15). Then they investigated the solidification paths of Mg–Al–Ca alloys with particular attention to the formation of the (Mg, Al)_{2}Ca phase, and argued that in the Mg-rich corner of the ternary phase diagram, the Al_{2}Ca phase cannot directly form from liquid during solidification, but forms by transformation of the (Mg, Al)_{2}Ca phase due to solid phase transformation [6].

However, in our experiments, it is suggested that the formation of Ca-containing phases is controlled by the Ca/Al mass ratio. In AZC912 with a Ca/Al ratio of 0.23, Al_{2}Ca and Mg_{17}Al_{12} coexist during the solidification process. In AMC502 with a Ca/Al ratio of 0.32, it consists completely in (Mg, Al)_{2}Ca, while in AMC503 with a Ca/Al ratio of 0.54, beside the formation of (Mg, Al)_{2}Ca, the Mg_{2}Ca phase appears. It seems therefore that further investigations are still required in this region of the Mg–Al–Ca ternary phase diagram.

The characteristic solidification reactions observed in this investigation are listed in Table 2 and the corresponding temperatures and solid fractions are listed in Table 3. These characteristic points are also summarized in Fig. 6, from which the solidification pathways of all alloys can be clearly described. It should be pointed out that some lines are drawn as broken ones in Fig. 6, which correspond to an uncertainty about their locations because...
of the limited number of experiments in our current study. From Fig. 6, the effects of Ca addition on the liquidiu and solidus temperatures of the various alloys can be clearly seen. An increase of the Ca content decreases the liquidiu temperature for both types of alloys. This result is consistent with the thermodynamic calculations of Djurdjevic and Schmid-Fetzer carried out for Mg–9Al–0.75Zn containing various amounts of Ca up to 5 wt% [24]. The effect of Ca on the solidus temperature is more complicated. In AZC91x alloys, the solidus temperatures first increases from 418 °C for AZ91 to 432 °C for AZC911 and then drops to 423 °C for the AZC912 alloy. The solidification of these alloys finishes with the formation of the non-equilibrium eutectic α-Mg + β-Mg17Al12, although the addition of Ca decreases the amount of β-Mg17Al12 phase. On the other hand, in AMC50x alloys, the solidification finishes with the non-equilibrium eutectic α-Mg + Mg17Al12 just like AZC91x alloy if the Ca content is smaller than 1%. When the Ca content reaches 1.89 wt%, the formation of Mg17Al12 is completely suppressed as observed in AMC502, the eutectic reaction involving α-Mg and (Mg, Al)2Ca at 514 °C, which means that the solidus temperature of this alloy is greatly increased compared to the AMC501 alloy in which it is 418 °C. In AMC503, the formation of Mg2Ca at 504 °C is the last step of the solidification. It should be mentioned that these maps shown in Fig. 6 are generated under non-equilibrium conditions, which is different from the equilibrium phase diagram [7,8]. In addition, regardless of non-equilibrium or equilibrium solidification, many factors, such as cooling rate, minor alloying elements, and even temperature acquisition system, may also have a great influence on the critical temperature values, so that transformation temperatures reported here may be somewhat different from those given by the equilibrium phase diagram [7,8] or in other reports [24]. Nevertheless, these maps can provide some useful information about the phase formation in alloys with different Ca levels and constitute a guideline for alloy selection.

It is reported that the introduction of Ca addition could increase the hot tearing susceptibility of AZ91 [25,26] and AM50 [1], but when the Ca level was higher than 2 wt% in AM50, the problem of hot tearing significantly reduced [1]. By re-examining the solidification characteristics of the Mg–Al–Ca alloys in Figs. 5 and 6, it is found that, the increase of Ca addition increases the formation temperature of Ca-containing phases and decreases the solid fraction at dendrite coherency point. Based on the feeding mechanisms for casting proposed by Campbell [27] and recently developed by Dahle et al. [28], lower fDCP value means that mass feeding finishes and interdendritic feeding starts to dominate at a higher liquid fraction, shrinkage being compensated by interdendrite flow of liquid at a very early stage during solidification. In AZC91x alloys and AMC501 alloy, the Al2Ca phase formed in the mushy zone after the dendrite coherency point can hinder interdendrite feeding, so that both of these mechanisms increase the tendency of hot tearing formation at the end of the solidification. On the other hand, in AMC502 and AMC503, although their fDCP is lower.

Table 2
List of the reactions identified during solidification by thermal analysis together with dendrite coherency point

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<thead>
<tr>
<th>ID#</th>
<th>Transformation</th>
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<tr>
<td>1</td>
<td>Start of solidification of α-Mg (liquids)</td>
</tr>
<tr>
<td>2</td>
<td>Maximum growth of primary α-Mg phase</td>
</tr>
<tr>
<td>3</td>
<td>Dendrite coherency point</td>
</tr>
<tr>
<td>3’</td>
<td>Eutectic reaction of α-Mg and Al2Ca.</td>
</tr>
<tr>
<td>3’</td>
<td>Eutectic reaction of α-Mg and (Mg, Al)2Ca.</td>
</tr>
<tr>
<td>4</td>
<td>Non-equilibrium eutectic of α-Mg and Mg17Al12</td>
</tr>
<tr>
<td>4’</td>
<td>Binary eutectic reaction of α-Mg and Mg2Ca (in AMC503 alloys)</td>
</tr>
<tr>
<td>5</td>
<td>End of solidification (solids)</td>
</tr>
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</table>

Table 3
Thermal analysis results for the temperatures (°C) and solid fractions (%) for at the critical points during solidification for all the alloys

<table>
<thead>
<tr>
<th>Alloys/reactions</th>
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<th>4 (4’)</th>
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</table>

Fig. 6. Solidification pathways of Mg–Al–Ca alloys: (a) for AZC91x and (b) for AMC50x. The broken lines mean some uncertainty for the border of the characteristic lines and regions.
than AM50 and AMC501, their solidus temperatures are much higher than that of the other alloys investigated in this paper, resulting in a narrower solidification interval; furthermore, no second phase forms between DCP and solidus, so that interdendrite feeding cannot be hindered until the very late stage of solidification in these alloys. Thus the hot tearing problem in the AMCSxx alloys with more than 2% Ca could be reduced to a large extent. In fact, the most successfully used commercial alloys AXJ530 and ACM522 are based on AMC502 and AMC503 [1]. With the above consideration in mind, it appears that, although the hot tearing susceptibility is influenced by many parameters, the solidification interval, the solid fraction at dendrite coherency and the formation of second phase in the mushy zone after dendrite coherency are closely related to hot tearing formation. This issue must be considered carefully as reported in many other alloy systems [29,30].

5. Summary

The solidification behavior of Mg–Al–Ca alloys (AZC91x and AMC50x alloys) has been investigated in some details by using thermal analysis and microstructural observations. The addition of Ca generates new Ca-containing phases and progressively suppresses the formation of the Mg17Al12 phase, the Ca-containing phases transform from Al2Ca to (Mg, Al)2Ca or Mg2Ca depending on the Ca content and the Ca/Al ratio. Moreover, the addition of Ca into Mg-Al alloys decreases the liquidus temperature, whereas its effect on the solidus temperature differs greatly from one alloy to another. In particular, the solidus temperature increases sharply in AMC502 and AMC503 owing to the absence of the Mg17Al12 phase which results in a significant reduction in their solidification interval. Ca addition also decreases the solid fraction at which dendrite coherency occurs in both alloys. It appears that the Al2Ca phase formed at temperatures between DCP and solidus may hinder interdendrite feeding, thus increasing the tendency of hot tearing formation at the end of the solidification.

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