Research on Porosity Defects of Al-Si Alloy Castings
Made with Permanent Mold

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Contents

Chapter 1 Introduction 1
1.1 Motivations and Background 1
1.2 The Current Status of Porosity Prediction 2
1.3 The Purpose of This Research 2

Chapter 2 Literature Review __ Overview of Porosity Formation 4
2.1 Definition of Porosity Defects for Al-alloys 4
2.2 Mechanism of Porosity Formation of Al-alloys 4
2.3 The Formation of a Gas Pore 4
   2.3.1 The critical condition to form a hydrogen gas pore 4
   2.3.2 Nucleation sites 5
2.4 The Volume Shrinkage, Inerdentritic Feeding and Porosity 6

Chapter 3 Literature Review __ Porosity Prediction for Al-alloy Castings 8
3.1 Modulus and Equisolidification Time Method 8
   3.1.1 Modulus method 8
   3.1.2 Equisolidification time method 9
   3.1.3 The deficiency of the modulus and equisolidification time method 10
3.2 Criterion Function Method 10
   3.2.1 Thermal parameters used for criterion function method 10
   3.2.2 Temperature gradient, G and other existing criteria 10
   3.2.3 The Niyama criterion 12
3.2.3.1 The popularity of the Niyama criterion
3.2.3.2 To apply the Niyama criterion to long freeze range (LFR) Al-alloys

3.3 Direct Numerical Simulation Method
3.3.1 Models considering only the density change due to solidification
3.3.2 Models considering both solidification shrinkage and gas evolution

Chapter 4 The Influences of Controlling Parameters in Foundry

4.1 Metal Constitutes
4.1.1 Grain refining effect of the three-minute elements, Ti, Zr and V
4.1.2 The influences of macro-structure on porosity
4.1.3 The influences of the minute elements on tensile and fatigue strength at an elevated temperature

4.2 Metal Quality
4.2.1 Purifying
4.2.2 Degassing
4.2.3 Quality check and controlling

4.3 Si-Refining
4.3.1 Eutectic Si refinement
4.3.2 Primary Si refining _ P inoculation
4.3.2.1 Mechanism of P Inoculation
4.3.2.2 Key points on P inoculation
4.3.2.3 P inoculation on porosity

4.4 Casting Processing
4.4.1 Mold designing
4.4.1.1 The gate ratio
4.4.1.2 The pouring basin
4.4.1.3 The sprue
4.4.1.4 The runner
4.4.1.5 A filter applied in the gate system
4.4.1.6 The gate
4.4.2 Foundry operation
4.4.2.1 Pouring temperature
4.4.2.2 Mold cooling
4.5 Inserts
4.5.1 Metals inserts (cast iron and steel)
4.5.2 Non-metal inserts (salt, fiber-reinforce material, and sand-core)

Chapter 5. Preliminary Calculations for Using Computer Simulation to Predict Porosity

5.1 The Criterion to Use — the Niyama Criterion, G/R^{1/2}
5.2 Things to Be Noticed While Using the Niyama Criterion
   5.2.1 The critical value of the Niyama criterion
   5.2.2 The moment to calculate the Niyama criterion
   5.2.3 The cooling rate used for calculating the Niyama criterion
   5.2.4 The influences of calculation conditions on the Niyama criterion values
      5.2.4.1 Element sizes
      5.2.4.2 The mold/casting interface heat resistance
      5.2.4.3 The Initial metal temperature
5.3 Correlation Between the Potential Porosity Location and the Niyama Criterion Values
Chapter 6. Reducing Porosity of Aluminum Permanent Mold Castings in Daily Production Aided by Simulation

6.1 Porosity Around a Non-aluminum Insert__Porosity at the Ring-carrier Area of a Gravity Al Piston
79

6.2 Porosity at a T-junction Area__Porosity at the Ingate Area of a Gravity Al Piston
86

6.2.1 The influences of the T-junction's structure
87

6.2.2 The Influence of the mold temperatures
90

6.3 Centerline Porosity and Porosity at a Hot Spot Area__Porosity at a Final Solidification Area of an Al Squeeze Casting
92

Chapter 7 Conclusions

7.1 Porosity Prediction
96

7.2 The Influences of Controlling Parameters in Foundry
97

7.3 Reduce Porosity of Al-alloy Permanent Casting Aided by Computer Simulation
97

Acknowledgements
99

List of Publications
100
Chapter 1 Introduction

1.1 Motivations and Background

The basic principles behind casting processes are straightforward. Molten metal of sufficiently low viscosity flows into cavities of shape complexity, and solidifies upon cooling. However, behind this simple principle lies many complicated reactions and phase transformations. If proper care is not taken, metal castings, in particular the aluminum alloys (Al-alloys), are prone to defects, such as porosity, one of the chronic problems, which impact the quality of the castings and worse the mechanical properties, such as tensile strength and fatigue life\textsuperscript{11, 12}.

Porosity forms when there is a gas entrapment, solidification shrinkage due to failure of inerdendritic feeding, and/or precipitation of dissolved gas from the molten metal. Inclusions also play an important role as they serve as nucleation sites for dissolved gas and thus facilitate gas pore formation. The effect of inerdendritic feeding is mainly influenced by the solidification pattern, i.e., columnar or equiaxed growth, which is decided by alloy constitutes and the casting process parameters. Hydrogen is the only gas dissolved to a significant extent in the melt of Al-alloys. It is, however, a constant source of difficulty for foundry-men because it dissolves upon reaction of molten metal with atmospheric humidity and the moisture of additions, such as constitute element ingots.

The task of a mold designer and foundry engineer is to make an optimized geometric casting design and choose proper process parameters that eliminate or minimize defects evolution while ensuring the product shape and structure. But porosity formation is a complex phenomenon where the final sizes and the distribution of porosity voids are determined by several strongly interacting process and alloys.
variables. As the result, it is usually difficult to eliminate porosity completely from Al-alloy castings, while reducing it or moving it to an unimportant area can be a choice.

1.2 The Current Status of Porosity Prediction

Until recently, the manufacturing of most Al-alloy castings was based on trial and error. Casting process parameters or casting design geometry were modified accordingly and the trial process would be interacted till desired product quality was achieved. This process can be tedious and time consuming.

Starting from the middle of 1980s’, due to the decreasing cost of computers and advances in computing methods, computer simulation of foundry process has been developed and improved by both academic and industry. Studies on porosity have then stepped forward from experiment-based investigations to computer simulation aided research. Most research jobs have been done to explore the mechanism of porosity formation and the ways predict it. There have been, however, very few publications whose results can be directly applied in mass production because the results of the studies have not been confirmed with tests in manufacturing scale.

1.3 The Purpose of This Research

With the purpose to figure out some useful countermeasures with which porosity defects can be reduced in mass production, a through literature survey on porosity covering the recent past thirty years has been made. In specific, the existing thermal-parameter based criteria to predict porosity have been reviewed. A summary regarding to the whole foundry process, starting from melting, metal treatment, and casting designing to process parameters controlling, based on the
author’s working and research experiences, has been made. Various simulation calculations performed with the purpose to reduce porosity defects in daily manufacture for some aluminum engine components have been reviewed. Most calculation results have been verified with confirming tests and applied to mass production. This thesis summarizes the jobs done with the expectation that it can be a useful guidebook to help foundry people who have been irritating by porosity defects while making Al-alloy castings with permanent mold.

Reference


Chapter 2  Literature Review __ Overview of Porosity Formation

2.1 Definition of Porosity Defects for Al-alloys
Undesirable voids in a solidified casting are called shrinkage or porosity defects according to their volume and the methods to detect them. Porosity is used to express dispersed pores that are in micro-scale and can only be detected by density measurement or microscopy. This kind of defects is often found in alloys with mushy solidification pattern, like Al-alloys.

2.2 Mechanism of Porosity Formation of Al-alloys
It is well accepted that porosity forms in Al-alloys due to the following reasons:
(1) The rejection of gas, mainly hydrogen, from the liquid metal because of the solubility changes during solidification;
(2) The inability of liquid metal to feed through the inerdendritic region to compensate for the volume shrinkage associated with the solidification.
Usually gas-pores form first, and shrinkage contributes to increase the dimensions of the voids. These voids, named porosity, appear most frequently in-between dendrite arms, and in most cases at the areas that solidify last.

2.3 The Formation of a Gas Pore
2.3.1 The critical condition to form a hydrogen gas pore
Since hydrogen (H₂) is the only gas dissolved to a significant extent in Al-alloys, the discussion in below will be concentrated on the formation of a H₂ gas pore. The pressure difference to form a gas-pore is described by following eq.¹)

\[ \Delta P = \frac{2 \rho}{r} \]  \hspace{1cm} (2 - 1)
where $\mathcal{W}$ is the interfacial energy of $\text{H}_2$ gas and the liquid metal, $r$ is the radius of the pore, and $\Delta P$ is the pressure difference between the exterior and interior of the pore. This eq. can be written as another form to show the critical radius of pore nucleation,

$$r^* = -2 \mathcal{W} / \Delta P$$  \hspace{1cm} (2 - 2)

Since pore growth is a diffusion-controlled process, the size of a pore is, therefore, influenced not only by the hydrogen content in the melt, but also by cooling rate during solidification. A higher cooling rate reduces the pore size by limiting the time of the pore growth.

2.3.2 Nucleation sites

(1) Homogeneous nucleation

The eq. (2.2) gives the critical size by which whether a nucleated pore will survive or disappear can be judged. An estimation using 2 atomic cross as the critical radius and surface tension for aluminum liquid metal gives a value for $\Delta P$ as 30000 atm. This reflects the real difficulty of homogeneous nucleation of pores in liquid metal. In practice, homogeneous nucleation is almost impossible because it needs such a big interior pressure.

(2) Heterogeneous nucleation

The difficulty of nucleation is reduced by the presence of surface-active impurities in liquid metal, since absolute pure liquid metal is impractical. Comparing with homogeneous nucleation, heterogeneous nucleation is easier by a factor

$$P_{\text{het}} / P_{\text{hom}} = 1.12 \left\{2 - \cos \mathcal{W} \right\} \left\{1 + \cos \mathcal{W} \right\}^2 / 4^{1/2}$$  \hspace{1cm} (2 - 3)

where $\mathcal{W}$ is the contact angle of the metal liquid with a solid (any of the impurities). $P_{\text{het}}$ and $P_{\text{hom}}$ are the interior pressure needed for heterogeneous nucleation and homogeneous nucleation of a $\text{H}_2$ pore, respectively. This eq. shows that a solid that
is completely wetted by the metal liquid (θ = 0, then the above factor is 1.12) is not a favorable site for a H₂ gas pore to nucleate on. On the other hand, a solid which is totally not wetted by the metal liquid (θ = 180, then the above factor is 0) is a good site for bubble nucleation. According to John Campbell, nucleation on solid does become favorable until the contact angle exceeds 60 or 70 degrees (Fig. 2.1)\(^2\).

![Figure 2.1 The pressure ratio of nucleating a gas pore heterogeneously/homogeneously Vs. the contact angle](image)

From this graph, it is clear that heterogeneous nucleation on the most non-wetted solid known (maximum θ is 160) requires only about one twentieth of the gas pressure as required for homogeneous nucleation in the bulk liquid.

### 2.4 Volume Shrinkage, Inerdendritic Feeding and Porosity

Most Al-alloys become denser in the phase change from liquid to solid. When the solidification is unidirectional, feeding of volume shrinkage is realized by the coming down of the liquid surface due to the gravitation effect in the earlier stage. Then metal moves through the inerdendritic channels driven by pressure drop due to
solidification. Porosity voids appear at the areas solidifies last when none of the feeding is successful.

For a casting with complex shape, when the solidification occurs due to 3-dimensional heat subtraction, several liquid pools, most frequently partial liquid and partial solid areas will form during solidification. For these isolated areas, inerdendritic feeding is the only driving force to move metal around. Theoretically speaking, shrinkage voids will form at these areas, only with the location changed according to the solidification order of the isolated areas themselves. It has to be mentioned that to nucleate a shrinkage void can be more difficult than to enucleate a \( \text{H}_2 \) gas pore by homogeneous nucleation. Because the interfacial energy between liquid metal and the air has to be overcome, and the motive energy like interior \( \text{H}_2 \) gas pressure does not exist. On the other hand, nucleating a shrinkage void on a gas pore can be as easy as nucleating a gas pore heterogeneously. As the result, shrinkage voids formed in any of the isolated liquid pool will usually not be just a pure air void, but a combination of shrinkage voids and gas pores. In other words, porosity formation in Al-alloys is the result of \( \text{H}_2 \) gas-pore formation and the failure of inerdendritic feeding to volume shrinkage.

Reference


Although the phenomenon of porosity formation has been well understood, the time to predict the defect precisely has not yet come. In the past fifty years, especially in the recent twenty years, research efforts have been made to predict porosity with the help of computer simulation. The studies made can be classified as the following three approaches:

(1) Modulus and equisolidification time method, which determines the areas that solidify last.

(2) Criteria function method, which calculates or regresses parameters to characterize resistance to inerdendritic feeding.

(3) Computer simulation method, which directly simulates the formation of porosity by mathematically modeling the solidification process.

The important results of these studies are reviewed in the below.

3.1 Modulus and Equisolidification Time Method

3.1.1 Modulus method

The modulus method is based on Chvorinov's rule\(^1\) that solidification time, \(t\) of a casting area is proportional to the square of its volume to area ratio, \(V/A\), named modulus.

\[ t = B (V/A)^2 \]  \hspace{1cm} (3 - 1)

\(B\) in this eq. is a factor that depends on the thermal properties of the metal and mold material. This experiment-based eq. has been testified by other researchers\(^2, 3, 4\), and was incorporated to some computer programs with which the solidification order
of a 2 or 3-dimensional model can be calculated\textsuperscript{5, 6} (Fig 3.1).

![Shrinkage prediction by Modulus Method](image)

Fig. 3.1 Shrinkage prediction by Modulus Method\textsuperscript{5}

3.1.2 Equisolidification time method

With the introduction of finite element/difference method to foundry field, equisolidification time contours or other isochronal contours could readily be calculated\textsuperscript{7, 8, 9}. The principles of the calculations are well established, and the results calculated are in good agreement with the corresponding experimental results in showing the last solidification area (Fig. 3.2).

![Porosity prediction of an engine block casting of Al-9.6%Si-3.8%Cu by equisolidification time method](image)

Fig. 3.2 Porosity prediction of an engine block casting of Al-9.6%Si-3.8%Cu by equisolidification time method\textsuperscript{9}
3.1.3 The deficiency of the modulus and equisolidification time method

To date, the determination of the areas that solidify last can be successfully carried on either by the modulus calculation or equisolidification time calculation based on numerical simulation of heat transfer. In estimating solidification sequence, the later is more accurate than the former, because modulus calculation does not take into account the mold temperature variation and the metal material physical properties. Therefore, the numerical simulation of heat transferring represents the most important application of computer simulation in foundry industry currently. But both methods have their limitation in predicting dispersed porosity, since they do not consider such factors, as interdendritic feeding and gas evolution, which govern separately or cooperatively the formation of dispersed porosity. This approach is, however, reliable in predicting gross shrinkage.

3.2 Criterion Function Method

3.2.1 Thermal parameters used for criterion function method

Due to the inefficiency of the modulus and equisolidification time method in predicting centerline and dispersed porosity, the criterion function approach has received considerable attention in porosity prediction. These criteria reflect the limiting conditions of interdendritic feeding. They are associated with thermal parameters, such as local temperature gradient $G$, cooling rate $R$, solidification velocity $V_s$ and local solidification time $t_f$. A combination of these parameters, which can be easily obtained from the numerical solutions of solidification heat transfer, is often applied.

3.2.2 Temperature gradient, $G$ and other existing criteria

The importance of temperature gradient was first proposed by Bishop et al\textsuperscript{[10]}, and
developed by Niyama et al.\(^{11}\) into a computer simulation method. This criterion gives information directly related to interdendritic flow. Therefore, it can predict centerline porosity more precisely than the equisolidification time method (Fig.3.3).

![Diagram](image)

(a) solidification time in min    (b) temperature gradient G in deg/cm

Fig. 3.3 Comparison of G and equisolidification time method in predicting gross shrinkage and centerline porosity of a steel casting (13Cr-5Ni)\(^{10}\)

The existing thermal parameter criteria proposed in literature so far, including temperature gradient G, are tabulated in Table 3.1.

<table>
<thead>
<tr>
<th>Criterion</th>
<th>Submitter</th>
<th>Time of publication</th>
</tr>
</thead>
<tbody>
<tr>
<td>G</td>
<td>Bishop et al.</td>
<td>1951</td>
</tr>
<tr>
<td>G/V(_s)</td>
<td>Davies</td>
<td>1975</td>
</tr>
<tr>
<td>1/V(_s)^n</td>
<td>Khan</td>
<td>1980</td>
</tr>
<tr>
<td>G/R(^{1/2})</td>
<td>Niyama et al.</td>
<td>1982</td>
</tr>
<tr>
<td>G/V(_s)</td>
<td>Lecomte-Beckers</td>
<td>1988</td>
</tr>
<tr>
<td>G(^{0.33}/V(_s)^{1.67})</td>
<td>Lee et al.</td>
<td>1990</td>
</tr>
<tr>
<td>G(^{0.38}/V(_s)^{1.62})</td>
<td>S.T.Kao et al.</td>
<td>1994</td>
</tr>
<tr>
<td>1/t(_s)^mV(_s)^n</td>
<td>F.Chiesa</td>
<td>1998</td>
</tr>
</tbody>
</table>

Nomenclature:
- G: temperature gradient
- V\(_s\): solidification velocity
- R: cooling rate
- t\(_s\): local solidification time

All these criteria can be reduced to the form of G\(^x\)/V\(_s\)^\(y\) (x varies over the range 0~2 and y varies over the range of 0.25~1), among which the Niyama criterion that can be reduced to G/V\(_s\) is a representative one.
3.2.3 The Niyama criterion

In 1982, Niyama et al.\textsuperscript{12) found that the critical temperature gradient was inversely proportional to the square root of the solidification time (Fig. 3.4). Therefore, they proposed to use $G/R^{1/2}$ at the end of solidification as a criterion for porosity prediction. This criterion was justified by Darcy's Law, so that it included the physics behind the difficulty of providing feed liquid in the last stages of solidification when the interdendritic liquid channels are almost closed. The critical value of the criterion was proven to be independent of casting size, first by Niyama et al. (Fig. 3.5), and later by other researchers (Table 3.2)\textsuperscript{13).}

Table 3.2 Proposed and calculated critical values of several solidification parameters for centerline porosity prediction\textsuperscript{13)}

<table>
<thead>
<tr>
<th>Parameters</th>
<th>Proposed Critical Values</th>
<th>Calculated Critical Values for Plate Thickness Listed</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>50mm</td>
</tr>
<tr>
<td>$G$</td>
<td>0.22 - 0.44</td>
<td>1.8 - 2.2</td>
</tr>
<tr>
<td>$G/R^{1/2}$</td>
<td>1.0</td>
<td>0.92 - 1.1</td>
</tr>
<tr>
<td>$P$</td>
<td>0.25</td>
<td>0.037 - 0.042</td>
</tr>
</tbody>
</table>

Fig. 3.4 The relation between the experimentally determined critical temperature gradient $G$ and the calculated solidification time $t_f$\textsuperscript{12)
Fig. 3.5 The relation between the experimentally determined critical Niyama criterion $G/R^{1/2}$ value and the calculated solidification time $t_{12}$.

3.2.3.1 The popularity of the Niyama criterion

This criterion has been widely integrated into current existing computer software to relate the output of the numerical heat transferring calculations (temperature gradient, solidification time, etc.) to empirical findings on porosity. The reason of its popularity can be attributed to the followings:

(1) The criterion itself is simple and only requires data obtainable from temperature measurements for verification.

(3) $G/R^{1/2} = (G/V_s)^{1/2}$, while $G/V_s$ is the most important parameter governing the constitutional undercooling, and hence decide the range of mushy zone, columnar or equiaxed growth in solidification. The critical condition of columnar growth is, $G/V_s \leq mc_0(1/k-1)/D$, in which $m$ is liquidus slope, $c_0$ is alloy composition, $k$ is the equilibrium distribution coefficient, and $D$ is diffusion coefficient in liquid. Therefore, this criterion has essentially a close relation with the solidification process, and hence porosity formation.

(4) The final solidification areas usually have a lower value of $G/R^{1/2}$, because these
areas usually has a lower G but higher $V_s$. The former is caused by the
deteriorated heat transferring condition at a final solidification area, while the later
occurs due to the phenomenon named as the acceleration of solidification \cite{18, 19}.

(5) The authors have proposed a critical value of $10 \, (\text{deg}^{1/2} \, \text{min}^{1/2} \, \text{cm}^{-1})$, and its
effectiveness has been verified with steel castings. There then exist different
values for different materials since the value is influenced by material properties
as declared by the authors.

3.2.3.2 To apply the Niyama criterion to long freeze range (LFR) Al-alloys

The discriminability of the Niyama criterion for casting steel has been well known.
The 1990s have seen a renewed interest in testifying whether this criterion is
efficient in predicting porosity for LFR aluminum alloys. Controversial opinions have
appeared.

(1) Opinions for the application

Laurent and Rigaut carried out some experiments with A356 alloy (Al-7Si-0.3Mg) cast
with sand mold, using risers of different sizes, with or without an end chill, grain
refinement, and modification under controlled hydrogen content. After comparing
the result of their experiments with that of the Niyama criterion, they found that the
minimum density (or maximum porosity) was located in the same position that
exhibited the minimum value of the Niyama criterion. They concluded that the
Niyama criterion was valuable when considering the relative values of the criterion
for a specific casting geometry and melt quality\cite{20}.

Huang and Berry used a statistical program to examine the correlation between
porosity in an Al-alloy (again A356, sand cast) and the several criteria functions\cite{21}.
They found that temperature gradient, $G$, and other criteria containing $G$, correlated
best with empirical data. Therefore, they argued that $G$ is the most influential
parameter among the criteria under study. They said that the critical temperature gradient to produce a porosity-free casting for both short freeze range (SFR) alloys and LFR alloys, depended upon the freezing time. This was just the reason why cooling rate was introduced to form the Niyama criterion\(^{11, 12}\). Their final conclusion was that thermal-parameter based criterion can be used to predict the relative porosity level for a casting of LFR alloys.

(2) The opinions against the application

Tynelius et al.\(^{22}\) formed a statistical model, which discussed quantitatively the influence of alloy types and processing conditions on porosity, based on experiments with A356 cast by both sand and permanent mold. The local solidification time, \(t_f\), and solidus velocity, \(V_s\) were said to be the most appropriate predictor for dispersed porosity among the parameters studied. Maximum pore size increased with an increase in local solidification time, \(t_f\). Concerning the area pore density, a longer \(t_f\) was beneficial for porosity formation at lower hydrogen content while a shorter \(t_f\) was beneficial at higher gas level. Increasing \(V_s\) made the threshold hydrogen content for porosity formation lower, i.e., a larger \(V_s\) is always beneficial for porosity formation. The authors of this research argued the suitability of the Niyama criterion to LFR aluminum alloys in considering the influence of \(t_f\) on maximum pore size and area pore density at a lower gas level.

Spittle et al.\(^{23}\) claimed that the Niyama criterion did not correlate with the pattern of the microporosity distribution in a simple cylindrical casting of Al7SiMg alloy, solidified progressively towards the feeder. They made further experiments with Al7SiMg plate castings having three different thickness (8, 15, and 25 mm)\(^{24}\), it was found that the highest values of porosity were associated with low \(G\), high \(V_s\), and long \(t_f\), in the case of the thicker plate (25 mm). They claimed that because of the
interdependence of the variables controlling percentage porosity in LFR alloys under multidirectional freezing conditions, none of the variables alone could be used to predict the locations having the highest porosity values. They suggested that extensive experiments under controlled directional freezing and multiple regressive analyses of data to obtain a criterion function for the prediction of microporosity in a given LFR alloy.

(3) Opinions for the application but with conditions

Overfelt et al.\(^{25}\) pointed out that the use of criteria functions derived from the physical description of inerdendritic flow was only likely to be effective when dissolved gas contents were low, and the presence of oxides and other porosity nucleating agents is minimized and when the overall solidification pattern from the casting through to the feeding system is truly progressive. They also emphasized the importance of selecting the appropriate thermophysical properties, particularly the conductance at the mold-metal interface in the calculations.

Viswanathan et al.\(^{26, 27}\) made temperature measurements and experimentally determined porosity distributions in grain refined Al-4.5% Cu alloys. They concluded that the criteria functions were dependent on casting conditions and alloy solidification mode. Casting processes and alloy types were categorized into four types with a different criterion selected for each type (Fig 3.6).

(a) For castings with SFR alloys or casting processes characterized by strongly directional heat removal (direct chilling or continuous casting), \(G/V_s\) (\(V_s\) was expressed with \(R\) in Fig. 3.6) calculated at the final stage of solidification is suitable.

(b) For castings with LFR alloys and high thermal conductivity, solidified in insulating molds, the temperature, \(T\) or solid fraction, \(f_s\) in the riser at the time when a particular location is at the final stage of solidification is suitable.
(c) For processes characterized by fine dendrite arm spacing and relatively high solidification velocities (permanent castings), the instantaneous cooling rate, $R$ (expressed as $\bar{R}$ in Fig. 3.6) at the final stage of solidification is suitable.

(d) For processes characterized by high temperature gradients and low solidification velocities (directional solidification processes), the suitable criterion is the same as for group (c).

As can be seen all the opinions, either for or against to use the Niyama criterion to LFR aluminum alloys, accept the importance of $G/Vs$, and it has been found that a shorter $t_i$ (higher $R$) is not always good in considering porosity density. When the hydrogen content is high, a shorter $t_i$ (higher $R$) is beneficial to porosity formation. This gives a possibility to apply the Niyama criterion to LFR aluminum alloys in industry where hydrogen content is well controlled.
3.3 Direct Numerical Simulation Method

Beginning from the 1980’s, there have been attempts to predict shrinkage and porosity quantitatively by direct numerical simulation. Same as in numerical simulation to calculate equisolidification time, a continuum is divided into infinite small convenient shapes, triangular or quadrilateral. These infinite small shapes are called elements. Efforts in this category can be classified into two groups: (a) considering only the density change due to solidification; (b) considering both solidification shrinkage and gas evolution.

3.3.1 Models considering only the density change due to solidification

The solid fraction was thought to be the key parameter in deciding the feeding mode for these models. Imafuku and Chijiwa calculated the shapes of shrinkage cavities based on this principle \(^{28}\). They categorized shrinkage defects as macroscopic and microscopic cavities, named shrinkage and porosity. The mechanism of the formation, and hence the predicting methods for the two classes of defects, were said to be different. The liquid flow induced by gravity was considered to be the cause of the formation of a shrinkage cavity in an isolated semi-solid region with low solid fraction, namely \(f_s < f_{sc}\), where \(f_s\) and \(f_{sc}\) represent solid fraction and the critical solid fraction, respectively. Solidification shrinkage in an isolated semi-solid region with high solid fraction, i.e. \(f_s > f_{sc}\), was considered to be the origin of porosity. They hypothesized that the lowest portion of the shrinkage cavity corresponded to the location of the disappearance of \(f_{sc}\) loop, and the location of the porosity corresponded to the region of the low solid fraction gradient when the \(f_{sc}\) loop disappears. The mathematical model they developed to calculate the volume of shrinkage cavity \(\Delta V_v\), generated during a time step \(t\) at an isolated non-solid region was,
\[ \Delta V_v = \int \beta \rho_s (f_s^* - f_{s,t}^* -(1 - v_{r,t}) - f_{s,t}^* - f_{s,t-\Delta t}^* (1 - v_{r,t-\Delta t})) dv \]  

(3 - 2)

in which \( v_{r,t}^* \) is the elemental shrinkage volume in an infinitesimal region of volume \( dv \) at time step \( t \), \( f_{s,t}^* \) is the solid fraction in \( dv \) at time step \( t \), \( \beta \) is the solidification shrinkage ratio, and \( \rho_l \) and \( \rho_s \) are the density of liquid and solid, respectively. The applicability of the mathematical model to practical problems was verified with steel castings \(^{29} \). One of the examples is given in Fig. 3.7.

![Fig. 3.7 Shrinkage prediction of a steel sand casting using eq. (3-2)\(^{28} \)](image)

A program to simulate shrinkage quantitatively was developed by Nagasaka et al in 1989\(^{30} \). In the program, a critical value of the solid fraction, \( f_{sc} \) was assumed to determine the liquid metal feeding mechanism, that is, liquid and mass feeding when \( f_s < f_{sc} \) but dendritic feeding when \( f_s > f_{sc} \). Macro-shrinkage and micro-shrinkage were predicted in regions where the solid fraction was lower and greater than the critical value, respectively. The solid fraction gradient, defined as \( f_s G_c \), was considered to be the key parameter that controlled the driving force for feeding in a high solid fraction region (\( f_s > f_{sc} \)). For steel castings with certain carbon content,
there existed a critical value of the solid fraction gradient, \( f_{scG_c} \), that was a function of \( f_s \) according to the authors. For instance, the relationship for steel castings with 0.4% C is,

\[
f_{scG_c} = 0.36f_s^2 - 0.36f_s + 0.09
\]  

(3 - 3)

When the maximum solid fraction gradient of an element is below the critical value, micro-shrinkage is expected at the element. On the other hand, mass feeding in the low solid fraction zone occurs easily by the force of gravity. The total shrinkage volume in this zone generates a macro-shrinkage cavity in the elements with a minimum pressure head, such as free surface elements. The basic concept with micro-shrinkage is similar to that of Imafuku et al with a different eq. for calculating of shrinkage volume.

\[
\Delta V_{s,i} = \beta \Delta f_{s,i} V_i = U_{ij} S_{ij} \Delta t
\]  

(3 - 4)

in which \( \Delta V_{s,i} \) is the shrinkage of element i during \( \Delta t \), \( \beta \) is the solidification shrinkage ratio, \( \Delta f_{s,i} \) is the increment of solid fraction for element i, \( V_i \) is the volume of element i, \( U_{ij} \) is the velocity between element i and its neighboring element j, \( S_{ij} \) is the area between element I and j, and \( \Delta t \) is the time step. It was said that good agreement between the calculated shrinkage value and the experimental results for steel castings had been obtained. A typical example from their results is shown in Fig. 3.8.

3.3.2 Models considering both solidification shrinkage and gas evolution

These models start to describe the behavior of inerdendritic flow. In 1966, Piwonka and Flemings introduced Darcy’s law, which is valid for fluid flow through a permeable material, into microporosity prediction\(^{31}\). They also evaluated the efficiency of inerdendritic feeding through the knowledge of local pressure drop.
Following this pioneering study, many researchers investigated the importance of inerndendritic feeding in microporosity formation by direct numerical simulation \cite{32, 33, 34, 35, 36, 37}. Some typical works are summarized in the below.

![Fig. 3.8 Calculated (with eq.3-4) and experimental shrinkage and porosity for a steel casting\cite{30}](image)

Ohnaka et al.\cite{34} tried to simulate the formation of shrinkage cavities in a complicated casting by modeling the motion of inerndendritic flow via solving the heat-and mass-conservation eq.s, and employing Scheil & Darcy's law. It was assumed that shrinkage cavities form in free surface elements, i.e. elements where the pressure is below a critical pressure and/or in which the pressure head is minimum. However, to use this model, accurate knowledge of permeability of the inerndendritic region, the boundary pressure for porosity formation, the critical solid fraction to calculate the amount of porosity, etc., are required. This limits its application. Besides, the model was two-dimensional at the time it was proposed.

Kubo and Pehlke\cite{35} developed a mathematical model based on the continuity eq.,
Darcy's law, and the conservation of energy and gas content. Their model suggested that the simultaneous occurrence of shrinkage and gas evolution was a key mechanism for porosity defect formation. Measured values of porosity in Al-4.5%Cu plate castings compared favorably with their calculated values (Fig. 3.9). They recommended minimization of gas content by degassing and increasing the mold chilling power for the production of sound castings. However, the recommended gas content and mold chilling power depend on casting shape, alloy composition, and required mechanical properties.

![Fig. 3.9 Porosity distribution of a 1.5cm-plate sand casting of Al-4.5% Cu](image)

Poirier et al.\textsuperscript{36} proposed a model to predict the formation and the amount of microporosity formed between primary dendritic arms of an aluminum alloy. Their model was similar to that of Kubo and Pehlke\textsuperscript{35}, with differences in calculating the permeability of inerpendritic flow and how the radii of gas bubbles depend on the sizes of the inerpendritic spaces. Their results indicated that porosity does not form when the gas pressure is below the pressure in the liquid, and porosity volume is proportional to primary dendrite arm spacing (Fig. 3.10).
A gas pore is stable provided that,

$$P_g - P = \frac{\Delta}{r_1 + 1/r_2}$$  

(3 - 5)

where $P_g$ is the pressure of hydrogen within the inerddendritic liquid, $P$ is the local pressure in the mushy zone, $\Delta$ is the surface tension of the liquid, and $r_1$, $r_2$ are the principle radii of curvature. According to eq. (3 - 5), it is easier for gas pores to form among primary dendrite arms than among secondary dendrite arms in a columnar mushy zone, because the spaces ($r_1$ and $r_2$) among the primary arms are larger than those among the secondary arms. They proposed the condition for porosity formation among primary dendrites arms as,

$$P_g - P = 4 \frac{\Delta}{g/da_1}$$  

(3 - 6)

Where $g_1$ is the local volume fraction of liquid, and $d_1$ is the primary dendrite arm spacing. The calculated value of porosity as a function of initial hydrogen concentration agreed well with the empirical data (Fig. 3.11). This graph also shows that an increase in temperature gradient ($G$) and solidification rate ($V_s$) results in less inerddendritic porosity.
Shivkumar et al.\textsuperscript{37} developed a mathematical model to simulate microstructure evolution and microporosity formation during the solidification of equiaxed structures for aluminum alloys. Parameters such as grain size, secondary dendrite arm space and eutectic space, together with the pore characteristics, such as the amount of porosity and pore size, can be obtained quantitatively with their model. It was found that both grain size and eutectic spacing varied inversely with cooling rate, a slower cooling rate resulted in a coarser secondary dendrite arm spacing because of the longer local solidification time. Their results indicated that as the cooling rate increased, there was a reduction in the total amount of porosity. But at cooling rates greater than 5° C/s, the amount of porosity is determined only by the hydrogen level (Fig. 3.12). The pore size decreased with an increase in cooling rate, and cooling rate has a greater influence on pore size than does the hydrogen content.
Among the three approaches described above, direct numerical simulation gives insight into the formation of dispersed porosity. But its application is mainly limited in research field for its complexity in use.

References

(6) N. Sirilertworakul, P. D. Webster and T. A. Dean, “Computer Prediction of Location


Chapter 4  The Influences of Controlling Parameters in Foundry

“Foundry is a black art! Foundry is a miracle!” These proverbs reflect the difficulty of foundry process because so many parameters need to be controlled at the same time and not all of them are controllable, as they are often interactive. Although foundry has a history as long as that of human beings, the typical foundry defects like porosity has never disappeared and will continue to trouble foundrymen as long as foundry operation continues.

In order to reduce foundry cost, foundry people are trying by all means to reduce porosity defects. Precious experiences have been accumulated in daily foundry operations. Good guideline textbooks can sometimes be written based on such kind of experiences. Aimed at to construct such a textbook for daily foundry of Al-alloy permanent mold castings, this chapter summarizes the important aspects of gravity permanent mold casting process based on the author’s working and research experiences.

4.1 Metal Constitutes

For most commercial alloys, the constitutes of the important elements are give in ranges, while only upper limits are given to those that are thought unimportant. For example, the constitutes of the commercial aluminum alloys AC8A is defined as $^1$)

\[
2.0 \sim 4.0\%Cu, \ 8.5 \sim 10.5\%Si, \ 0.5 \sim 1.5\%Mg, \ \leq 10\%Fe, \ \leq 0.2\%Ti, \ \ldots \text{etc.}
\]

Except for Cu, Si and Mg, all the other elements are only given with an upper limit, among which Ti is specified as less than 0.2%. There is no problem with such a specification when the other minute elements, such as Zr, V, do not present a noticing level. With the existence of Zr and V, a small difference in Ti content, both
under 0.2%, may cause a totally different macrostructure, even if all the other foundry parameters are the same.

4.1.1 Grain refining effect of the three-minute elements, Ti, Zr and V

The grain refining effect of Ti to aluminum alloys was realized from 1970s 2). The refining effect is caused by peritectic reaction between Ti and Al. Therefore, other elements like Zr, V, etc., that can have similar peritectic reaction would also have grain refining effect. Such an effect of Zr was reported in 1980s 3). The mechanism is theoretically a nucleation phenomenon. The metastable phase $\text{Al}_3\text{Ti}$, or $\text{Al}_3\text{Zr}$, formed from the peritectic reaction between Al and Ti, Al and Zr, respectively, behave as the nuclei of alpha Al. It needs to be pointed out that once the intermetallic compounds become stable, they lose the effect of refinement 4).

The use of the three elements, Ti, Zr and V in combination, in concentration of 0.1–0.4%, as addition to Al-Si alloys for pistons and cylinder heads, was proposed by a French patent 5). The main benefit claimed was the improvement in creep resistance at elevated temperature. Industries started to look for the best combination for the three elements, although the refining mechanism of V was still not clear. A range for each of the three elements started being given for some of industrial self-made alloys.

According to the author’s experience, a small change in the amount of the three elements can cause a significant change in the macrostructure of Al-alloy castings. Fig. 4.1 shows the macrostructures of an Al-Si alloy piston cast with exactly the same permanent mold and the same conditions, but a small differences in the three minute elements as shown below the pictures. The macrostructure changes from partial columnar and partial equiaxed (a) to whole equiaxed (b) (Fig. 4.1).
4.1.2 The influences of macro-structure on porosity

Different macrostructures are the results of different solidification patterns. Therefore, it has a correlation with porosity formation. In the author’s early research on columnar-equiaxed transition with cylindrical ingots, those ingots that contained the largest area of equiaxed grains also contained the most porosity. Conversely, those ingots contained none or few equiaxed grains contained the least porosity as well. Both dye-check and microstructure examination of casting (a) and (b) showed that the columnar + equiaxed structure presented no detectable porosity, while the whole equiaxed structure presented well-distributed fine porosity (Fig. 4.2).

Fig. 4.1 The effect of the three-minute elements on macrostructure

Fig. 4.2 Microstructure of area A taken from (b) of Fig. 4.1
To confirm the influence of macrostructure fineness on porosity formation, examinations have been done to another Al-Si piston, which has almost the whole equiaxed structure, but different in macrostructure fineness due to constitutional differences. The result was that the finer macrostructure contained more porosity voids in various dimensions than the coarser macrostructure (Fig. 4.3). A conclusion then can be made that the finer the macrostructure, the more porosity voids will form. Worth to mention that the data showing in Fig. 4.3 were taken from 5 piston of a new Al-Si alloy with high Cu and Ni, named M174. The practical constitutes of the alloy cannot be opened here for business reason.

According to the author's experiences, it is suggested to control the combined concentration level of Ti, Zr and V below 0.30% for any high cupper Al-alloy casting.
made with permanent mold. Since the total amount of the three minute elements has a magic effect in controlling macrostructure, and hence porosity, a minor adjustment in one of, or the three minute amount elements can help solve the well-distributed porosity problem readily sometimes. Nowadays, spectrum analysis has brought about a great convenience to daily routine chemical constitute analysis. It is therefore important to do the spectrum analysis regularly in daily production, and perform extra checks whenever an abnormal macrostructure, like that as shown in (b) of Fig. 4.1, is observed.

4.1.3 The influences of the minute elements on tensile and fatigue strength at an elevated temperature

With the condition that there is no porosity defects, the mechanical properties of the casting can expect an improvement. The laboratory tests in the author’s company have also confirmed the improving effects on tensile and fatigue strength at about 350°C with the addition of proper amount of the three elements. Whether the reasons of the strengthening effects come from the grain refinement or due to the well-distributed thermally stable intermetallic compounds, which behave as grain refiner in their unstable stage, needs further investigations.

4.2 Metal Quality

As often heard from foundrymen, “there would be no way to get a casting free from porosity, if the metal is not clean!” Good metal quality here means appropriate chemical compositions and a low gas content. The former can be checked by spectrum analysis, while the later need extra examination.

4.2.1 Purifying

In order to remove undesired metal constitutes, such as K and Na, and the organic
materials mixed from using machinery chips as the source material, Cl-containing chemicals, such as C₂Cl₆, is often used as purifying agent. But this operation is not allowed in the advanced countries like Japan because of the environment regulations. Then, countermeasures, such as double filter system, enough calming time, etc. should be applied in metal transferring before pouring, if alloying is being performed inside. When outside ingot providers are used, strict limits should be given to impurity constitutes when making the purchasing specifications.

4.2.2 Degassing

As mentioned in chapter 2 (see section 2.3.1), H₂ is the only gas dissolved to a significant amount in the melt of Al-alloys. However, it has a constant source because it is derived upon reaction of molten metal with atmospheric humidity and moisture of contained by whatever the molten metal contacts to.

\[ \text{Al} + \text{H}_2\text{O} \rightarrow \text{Al}_2\text{O}_3 + \text{H}_2 \]  
(4 - 1)

(1) Way of degassing and the agents used

Electronic Bubble Flow (EBF) with inert gas, such as Ar or/and N₂ are widely applied in industry to remove H₂ off from the melt. The degassing effect of Ar and N₂ are different, even though both of them are inert gases. A proportion decided from both cost and effect is often applied in industries. If allowed, a certain proportion of Cl₂ can be mixed into the bubbled gas, so that the purification process can be omitted.

(2) Frequency of degassing

The absorption of H₂ occurs continuously. As shown in Fig. 4.4, the hydrogen content will exceed 0.25cc/g after 7 hours even if there is no touch to the metal. Therefore, degassing operation should be performed with a regular interval. The interval depends on the maximum limit set to the melt and the humidity of the air,
while the later depends on seasons. In the rainy season, shorter intervals, or even continuous degassing is often applied. It should be noticed that this operation also causes continuous oxidization of fresh metal, though there is no worry for an over degassing.

4.2.3 Quality check and controlling

The Gas content can be easily measured by the Initial Bubble Method operated under a reduced pressure. Re-degassing is performed if the gas content is over the limit set beforehand. The cleanliness of the metal, however, can only be checked through measuring the density of the metal, because only density is the comprehensive reflection of porosity level. A convenient method, named Density Index method, which comparing the density of the sample solidified under a reduce pressure, namely $\rho_{\text{vac}}$, with the density of the sample solidified in the air, namely $\rho_{\text{air}}$, with the following equation,

$$\text{Density Index} = \left( \frac{\rho_{\text{air}} - \rho_{\text{vac}}}{\rho_{\text{air}}} \right) \times 100 \quad (4 - 2)$$

It is obvious that the index value will be zero, if the two densities are the same, which mean no porosity form under the vacuum condition. The lower the index value, the better the metal quality, while a upper limit, 1.2 is often proposed for Al-Si.
alloys. The samples for index density check are taken from the holding furnace right before pouring. And the values set for different alloys are different in daily production, depending on whether the alloy has a larger/smaller trend to form porosity.

4.3 Si-Refining

Si is a very bristle phase as being well known. Only when it is well distributed in the alpha matrix, can Si optimize the mechanical property of Al as an alloying element. Otherwise it behaves as a defect to Al alloys just as carban does to casting iron.

4.3.1 Eutectic Si refinement

Not until Na - modification technique was patented in 1921, did hypoeutectic Al-Si alloys come into commercial importance. Later on, Sr, Sb, etc., has been found to also have a similar refining effect on eutectic Si. This refinement dramatically enhances the alloys’ machinability and mechanical properties. The modification mechanism has been explained with various theories, in which attention has been mainly paid to the change of interfacial energy between eutectic alpha Al and Si, and between Si and the melt. It is found that there is a reduction in the interfacial energy mentioned above upon the addition of the refining elements, though a widely accepted theory has yet to be established.

The reduction in wear resistance after eutectic Si modification, the aggravation of high-temperature toughness, the difficulties of this operation in foundry, such as introducing gases to the melt and the fading phenomena of the refining effect, etc., have limited the application of eutectic Si modification to the alloys of piston and cylinder heads. Except for very small motorcycle pistons, which solidify at a higher
cooling rate and work at a lower temperature environment, eutectic modification is no longer being applied in piston manufacture.

4.3.2 Primary Si refining _ P inoculation

Before the finding that P inoculation could control the size and shape of primary Si, hypereutectic Al-Si alloy had not received serious attention because of their brittleness and lack of machinability caused by the large and irregular natural shape of primary Si crystals. The P inoculation is, however, applied not only to hypereutectic Al-Si alloys, but also to eutectic Al-Si alloys, because primary Si can frequently be seen for most industrial eutectic Al-Si alloys, especially at a thicker area where the cooling rate is low. This phenomenon is explained by coupled zone theory of Al-Si alloys\textsuperscript{10).} Just as the microstructure consisting of only eutectic can be obtained not only at exact eutectic composition, primary Si can also be obtained at the eutectic or even a hypoeutectic composition. Of course, other alloying elements can also have their contribution in altering the eutectic composition.

4.3.2.1 Mechanism of P Inoculation

P reportedly combines with Al in the melt to form tiny insoluble aluminum phosphide (AlP) particles that, due to their close crystallographic lattice constant to Si, acts as suitable nuclei on which primary Si grows during solidification. Both AlP and Si have a diamond cubic crystal habit and a similar lattice constant (Si, $a_0 = 5.43\text{Å}$; AlP, $a_0 = 5.45\text{Å}$). Although this theory has never been proven conclusively, a reduction in primary Si size by 90-92% with P inoculation has been observed, and the microprobe analysis has shown that the seeds of primary Si contain both Al and P\textsuperscript{11).}

Worthy to mention an unexpected finding, obtained when confirming whether P has refining effect on eutectic Si or not, that P alters the normal Al-Si eutectic composition toward a lower Si level, thereby causing primary Si crystals in 11.5%Si
melt\textsuperscript{12}, which would normally contain none of primary Si (refer to (a) and (b) of Fig. 4.5). From this picture, it can also be seen that P has no effect on the size or shape of eutectic Si, although there have been some argues on this.

![Fig. 4.5 The influence of P on eutectic composition 12)](image)

4.3.2.2 Key points on P inoculation

Having known the mechanism of P inoculation, there seem no hard rules applying to this operation. However, in order to obtain the expected effect, there are some important things need attentions. With years of experiences in using P to Al-Si alloys, the author considers the following aspects are to be understood and remembered.

(1) What to add?

Eutectic P-Cu containing 7-8\%P, instead of 15\%P-Cu, dissolves readily in Al-Si alloy melts and provides consistent and reliable refinement. For using in large-scale production, “shot” form is suggested because it is much cheaper than “brazing rod”.

(2) When to add?
Two questions are raised here: at what temperature and how long before pouring P should be added.

A higher temperature is preferred to obtain a better dispersion of AlP. But excess superheat causes a higher mean mold temperature that slows down the solidification and hence worse the refining effect. If the mold can be intensively cooled, this will not become a problem. But the bad effects, such as hydrogen pickup, melt oxidation and Mg burnout under a higher melt temperature should never be forgotten. The P addition temperature should generally be the desired pouring temperature decided based on other practical considerations such as feeding in coping with the practical mold temperature. No excess superheat is needed, but never add P at a temperature lower than 705 °C in order to get a good distribution of AlP.

The refining agent should be added to the melt after the melt has been otherwise prepared (degassed for example). This is because any agitation of the melt will cause agglomeration of AlP particles, flux gas bubbles also “float” AlP particles to the melt surface where they are entrapped with dross and removed during skimming. If P-Cu is added in the alloying stage, a P bearing salt, such as PCl₅, is suggested to be added just before pouring in order to provide active AlP for inoculation. Action like calming to remove gas introduced by P inoculation is always suggested.

(3) How long the refining effect retains?

The effect of P inoculation is not permanent. The refining effect is loosing not only due to the loss of P from the melt, but more due to agglomeration of the AlP particles. In other words, chemical analysis may give the same P percent after complete loss of refining effect as that during the period of effective refinement. But cluster of AlP can be detected by high magnification inspection.
The useful life of a refinement treatment is dependent on melt’s lot-weight and the stirring or agitating frequency that the melt undergoes. Under otherwise similar conditions, refinement remains effective longer in large melts than in small ones. For example, melts in the range of 2268kg-4536kg retained their refinement completely for 24 hours, while melts in the range of 91-227kg, refined with the same agent, retained refinement for only 4 to 5 hours\(^{12}\). Therefore, each company should do some experiment in order to find the proper holding time for their holding furnace volumes and operating condition.

(4) How much to add

The level of P needed in a melt to provide good refinement is quite small. It has been observed that 10 ppm (0.001%) can give adequate refinement, while 15ppm is considered as a safe level. However, because only 5-15% of the added P remains in the melt, an addition of at least 200ppm (0.02%) is generally required to reach the desired retained minimum. When P-Cu of 8%P is used, adding 0.25-0.3% of the melt P-Cu will release 0.02-0.024%P to the melt, of which 0.001-0.0036%P (10-37.5ppm) will be retained, which is adequate for inoculation.

(5) What happens if too much P is added?

According to the literature, there is no obvious harm regarding to Si size from excess P. However, excess AlP can be detrimental in another way while AlP behaves as an inclusion.

The first observation of massive AlP inclusion the author has experienced was during machining a batch of B390 (17%Si) alloy cast with permanent mold. What first appeared as darkened areas flush on the machined casting surface, expanded overnight to stand well proud of the casting surface. Unlike typical oxide-dross, this inclusion is very hard and has a ceramic-looking structure (Fig. 4.6). Various
attempts to identify the inclusion were made in the beginning, but not conclusive. EPA analysis has shown the inclusion containing 9.2 – 20% P, about 30% O, and other elements, except for Al.

A review of the melt handling and casting process of these parts were made to see what had been changed. It was noticed that 15% P-Cu was added to adjust Cu content of the alloy for the trouble batch castings. Inadvertently, P content was increased from 97ppm, at which there had been no such a problem, to 439ppm. By referring to similar inclusion in the literature \(^{13}\), the author considered that the problem had originated from AlP and the swelling of the darkened areas was some kinds of reactions occurring between AlP and moisture in the air or machining lube upon exposure during machining, for instance,

\[
\text{AlP} + \text{H}_2\text{O} \rightarrow \text{Al}_2\text{O}_3 + \text{P}_2\text{O}_5 + \text{H}_2\text{O} \quad (4 - 3)
\]

Then, the P-Cu added was replaced by pure Cu ingot for Cu-content adjustment. The defect disappeared immediately after the replacement. The problem has been admirably solved. From then, people working in the plant are more careful about P
amount in Al-Si alloys. A rule was made that never add more than 400ppm P to any of the Al-Si alloys.

4.3.2.3 P inoculation on porosity

The strengthening effect of P inoculation on the mechanical properties of Al-Si, due to the refining effects on primary Si, has been well established. But there has so far no report on the influence of P inoculation to porosity formation. Considering the improved condition of interdendritic feeding after the shape of primary Si modified, P inoculation should benefit porosity reduction. This effect has once been verified with a high-Cu (2.5 - 4%), Al-Si eutectic alloy piston cast with permanent mold. In the beginning, porosity on the finished machined surface of the land area of a piston, where there is a diameter change, can be recognized with dye-check (Fig. 4.7).

![Dye-check of the finished surface](image1)

(a) dye-check of the finished surface

![Microstructure of the red area](image2)

(b) microstructure of the red area (×100)

Fig. 4.7 Porosity at the land area of a piston

Various tries, such as intensive local mold cooling, have been tried. The situation became better but not conclusive. Finally a small amount of P bearing salt (5-10%P, 10-15%K, the rest Cl) was added before pouring and the porosity problem has been solved (not detectable with dye-check), though the real mechanism of the
improvement needs further confirmation.

4.4 Casting Processing

Casting processing consists of mold designing and casting processing-parameter specification. Comparing with the big progress and great efforts made with computer simulation in the past twenty years, development in casting-designing rules is so unobvious, as if it has been neglected for many years. Good textbooks containing some useful casting designing rules, such as the one written by Wlodawer 14), were published many years ago. It is not the case that there has been no progress in casting-designing rules development, but it is an area left consciously or unconsciously for the foundrymen to summarize themselves. Nevertheless, for objective or/and subjective reasons, there exists a big black in this area. This section serves to fill some part of the blank.

4.4.1 Mold designing

While making a mold design for a gravity casting, at least two things have to be taken into consideration. One is a desired mold filling to eliminating air and oxide-film entrapments in melt; the other is unidirectional solidification to avoid serious shrinkage and porosity in castings. In recent years, mold-cooling technology to improve productivity, which has long been applied to pressurized die-casting, has also been introduced to gravity permanent casting. With this introduction, the solidification order naturally decided by the casting structure has collapsed. The intensive mold cooling can be a useful way to remove/reduce porosity, and it can also cause porosity that may not occur if without it.

A bad gate system causes not only the random oxide-film inclusion, but also the random porosity. The oxide-film inclusions are created by the surface folding during
mold filling. The existence of oxide-film helps porosity formation, because it usually contains air, and itself is the good site for pore nucleation. A good gating system is tolerant of wide variations in foundry practice. Thus pouring will be under the control of the gating system, not the caster. Whether a gating system is good or not is decided by every part of the gating system, which usually are made of a pouring basin, a sprue, a runner, and a gate/gates. The designing of these parts is inter-related, because a proper area ratio of them, namely gate ratio, is required. Fig. 4.8 gives the image of a gating system frequently applied to a cylindrical castings cast with permanent mold.

Fig. 4.8 A gate system with vertical runner and two vertical gates

4.4.1.1 The gate ratio

Gate ratio is the sectional - area ratio of sprue, runner and the gate. There are unpressurized gating system and pressurized gating system. The former is an expanding system that slows the metal flow at each stage prior to entering the mold, while the later is a contracting system that choking the metal flow at one place and
causes the gating system to fill back from the choking point. It is a common sense to apply an unpressurized gate system to aluminum alloys. In literatures, there are some recommendations on the gate ratio for the unpressurized gate system: \( S_s : S_r : S_g = 1:1.2 \sim 2.2:1 \sim 4 \)\(^1\), in which \( S_s, S_r, \) and \( S_g \) represent the sectional areas of the sprue, runner and gate respectively. For a gating system with vertical runner and gates as shown with Fig. 4.8, the common area of the sprue and the runner, \( S_c \) in Fig. 4.8 is more important than the runner sectional area, \( S_r \) because \( S_r \) is not fully filled until the last stage of the mold filling. While making a gating system design for an aluminum casting, instead of trying to satisfy such a uncertain ratio, the gating system should be made to slow down the metal flow gradually, and to create as much opportunity as possible for the melt to become quiescent before entering the cavity. For example, even \( S_g = 2 S_s \) in Fig. 4.8, it is not a guarantee of a good gate system, if the gate is a high narrow slot because metal will then splash into the mold as a jet, and surface turbulence will impair the quality of the casting. When talking about gating ratio, the famous foundry expert John Campbell said “designing a gate system based on gate ratios is a mistake”\(^{15} \). The author also does not suggest paying much attention to the practical value of the gate ratio, but would like to emphasize one thing that, for the gating system with a vertical runner and gate/gates, the gate sectional area should be made at least the same as or even bigger than the runner sectional area as shown in Fig. 4.8, so that metal is slowed down just before entering the cavity.

4.4.1.2 The pouring basin

Pouring basin is the entrance through which molten metal is introduced into the cavity. According to the author’s experiences, following aspects are to be considered while designing a pouring basin. (a) The basin itself has to be filled up as
quickly as possible in order to reduce the pouring head. This can be realized by properly chocking of the sprue bottom, such as applying a filter at the sprue base like showing in Fig.4.8. (b) Melt should be poured from the blind end of the basin so that the fall of the stream is arrested, and bubbles and dross will have chance to float up to the surface. (c) The exit of the basin should be larger than the entrance of the sprue to avoid air aspiration at the basin/sprue connecting location. (d) The ideal profile of a pouring basin is the one that will introduce the melt to flow along the sprue wall, rather than make it flow towards or collide onto the wall, which causes collapse and entrapment of surface oxide film. Besides, a pouring basin with a round bottom will cause the tendency of the metal to rotate, forming a vortex and hence aspirating air. Fig. 4.9 shows two types of pouring basins used in our production. The type showing with (a) gave a higher rejection of random inclusions and porosity than that given by the type showing with (b), when all the other foundry conditions were same.

![Two types of pouring basin](image)

(a) the old type   (b) the optimized type

Fig. 4.9 Two types of pouring basin

4.4.1.3 The sprue
Sprue is the down-runner that introduces melt to the runner and provides driving force for mold filling in gravity casting. To design a sprue with right size is the key point in gating system design. An oversized sprue is, in particular, a liability because it leads to oxide built-up and air entrapment. How then to design a sprue with the right size?

(1) The height of the sprue

Theoretically, if a stream of liquid is allowed to fall freely from a starting velocity of zero, after falling a distance of h, it will reach its maximum velocity given by Bernoulli’s relation,

\[ V_{\text{max}} = \sqrt{2gh} \]  \hspace{1cm} (4 - 4)

in which \( \mu \) is the friction coefficient of the sprue-wall. In a gating system, the maximum velocity, \( V_{\text{max}} \) usually appears at the sprue exit. To keep laminar flow, the maximum velocity should be below a certain value theoretically decided by the Weber number \( W_e \).

\[ W_e = \frac{V_{\text{max}}^2 \mu}{\rho r} \]  \hspace{1cm} (4 - 5)

Where \( \rho \) is the density of the metal, \( \sigma \) is the surface tension and \( r \) is the radius of the surface curvature. Surface turbulence will occur if the \( W_e \) number exceeds its critical value. John Cambell has proposed 20% loss of friction, i.e., \( \mu = 0.8 \); and a limit range of 0.2-0.8 for \( W_e \) to avoid surface turbulence. Taking the upper limit of \( W_e \), for a thickness of 10mm filled with pure aluminum liquid, the \( V_{\text{max}} \) to keep the metal free from surface turbulence as calculated by eq. 4-5 will be,

\[ V_{\text{max}} = \frac{W_e \mu}{\rho r} = \frac{(0.8 \times 0.9/2500 \times 0.005)^{1/2}}{24 \text{ cm/s}} \]

And the corresponding head limit, \( h_{\text{max}} \) as calculated by eq. 4-4 will be,

\[ h_{\text{max}} = V_{\text{max}}^2/2g \sigma^2 = 24 \times 24/2 \times 980 \times 0.8^2 = 0.46 \text{ cm} = 4.6 \text{ mm} \]

However, a sprue of 4.6mm high is never impractical. This calculation tells us that
surface turbulence is actually occurring at the bottom of all the practical sprue in use. What we can do while making a gate system designing is trying to make the head, \( h \), as low as possible. This is particularly important when deciding the location of a vertical gate. The lower end of the gate should not be 5mm higher than the cavity bottom line, because there is no any brake, a filter for example, to slow down the metal there afterwards.

For gravity casting, a sprue should, at least, have a height not lower than the casting height in order to make filling driven by gravity possible. In practice, the sprue height, \( h_s \), for a gate system as shown in Fig. 4.8 is decided according to the following relation,

\[
h_s = h_c + h_r - h_p \quad (4-6)
\]

In which \( h_c \), \( h_r \), and \( h_p \) are the height of the casting, riser, and the pouring basin, respectively. The riser height, \( h_r \) is decided according to the riser amount needed after its horizontal dimension decided based on the feeding-distance principle. The height of a pouring basin, \( h_p \), should always be taken into consideration together with the sprue height, \( h_s \).

(2) The sectional area of the sprue _ the average flow rate of mold filling

The casting weight filled per unit time is called average flow rate. While a casting is made, the average flow rate of the casting can be easily obtained by simply dividing the casting weight, \( W_{\text{casting}} \), with the measured filling time, \( t \).

\[
Q_{\text{ave}} = \frac{W_{\text{casting}}}{t} \quad (4-7)
\]

For an unpressurized gate system as shown in Fig. 4.8, it is the exit area of the sprue, \( S_s \), that decides the average flow rate, because this area is the narrowest area in the gate system and it is supposed to be fully filled during pouring. This is why the flow rate topic is discussed here together with the sprue designing. When making a
gate system designing, the target average flow rate can be estimated by the following eq.,
\[ Q_{avg} = \frac{\rho \cdot V_{max}}{\mu} \cdot A_{exit} \]  
(4 - 8)

In which \( \rho \) is a coefficient reflect the effect of the filter, and friction of mold-walls, and whatever other parameters that give barrier to the metal. \( \rho \) is the density of the metal, \( V_{max} \) is the velocity of the metal calculated by eq. 4-4, and \( A_{exit} \) is the sprue exit area.

An extremely big flow rate is not good whether it is due to a high velocity or a large sprue exit area, because the former causes severe surface turbulence and the later offers difficulty for itself being fully filled, and hence accompanying with air aspiration and metal oxidization. On the other hand, if the flow rate is too small, some part of the casting will solidify, and cold-lap defects, such as cold-shut, will occur. The principle is that the filling time, decided by the flow rate, should be not longer than the time needed to solidify the thinnest section of the casting, even if some solidification be allowed during mold filling. However, it is worthy to note there is actually no lower limit, i.e., no “not shorter than” for the mold filling time.

For a specific casting, an average flow rate that seems reasonable according to long-time experiences can be used in the first-time gate system designing, and can always be modified in subsequent trials. The designed flow rate can be checked with a stopwatch when the mold is first poured.

Assuming the average flow rate, \( Q_{avg} \) has been known from the long-time experiences, how can we achieve it through gating system designing? A sprue has two areas, the entrance (top) and exit (bottom) area, \( A_{top} \) and \( A_{exit} \). From eq. 4 – 8,
\[ A_{exit} = \frac{Q_{avg}}{\rho \cdot \mu \cdot V_{max}} \]  
(4 - 9)

The sprue entrance area deducted from Bernoulli’s relation and the continuous eq.
then is,

\[ A_{Stop} = A_{Exit} (1 + hs/hp)^{1/2} \]  \hspace{1cm} (4 - 10)

For instance, if we wish to pour an aluminum alloy casting whose density is 2.72g/cm\(^3\) at an average flow rate of 500g/s, and the casting together with its riser height dictates the pouring basin and sprue height to be 5.5cm and 9.5cm, respectively (h = 15cm), \( V_{max} \) calculated by eq. 4-3 will be 137cm/s, and the sprue exit and entrance area calculated by eq. 4-9 and 4-10 will be 2.16cm\(^2\) and 3.57cm\(^2\), respectively. When the section shape as shown in Fig. 4.10 (a) is applied for the sprue, if the dimension A is set as 2.6cm, the dimension B calculated will be 0.8cm. And a 2 degrees taper to the back side of the sprue (ref. to Fig. 4.8) will make the sprue entrance area almost equal to the value calculated by eq. 4-10.

Whenever a new mold is designed, the sprue exit area should be calculated from eq. (4 - 9) due to its importance. But the sprue entrance area can be decided by a proper taper, 2\(\rightarrow\)3 degrees for instance, to avoid tedious calculation every time, especially when a specific cross-section is applied for the sprue.

(3) The shape of the sprue cross-section

Theoretically, if the sprue entrance and exit area are correctly sized, it does not matter whether the sprue cross-section is round or square, since the sprue is fully filled from the early stage. However, some authorities strongly suggest not applying round shape to avoid vortex\(^{15}\). Therefore, the shape of (a) in Fig. 4.10 is recommended while (b) and (c) are not.

To summarize, there is no substitute for a proper sprue design in gate system designing. A good sprue should be a vertical tapering channel, and the mold-filling
rate should be under the absolute control of the sprue exit area. If a casting is found in practice being filled too fast or too slow, the adjustment the sprue exit area should firstly be considered.

4.4.1.4 The runner

For an unpressurized gate system, an ideal runner should have a section-area that is at least twice the sprue exit and half the gate area, i.e., gate ratio = 1:2:4. A runner can be either horizontal or vertical depending on the orientation of the gate. A vertical runner is often applied when the gate is vertical as shown in Fig. 4.8. In this case, the gate area is usually determined by the casting parameters such as, casting height, locations of inserts, local casting thickness at which the gate is connected, etc. This pre-determined gate area could seldom be 4 times of the sprue exit. In other words, it is difficult to realize the gate ratio, 1:2:4. For example, the gate area in Fig. 4.8 dictated by the casting is only 2 times of the sprue exit area. Then, if the runner area has to satisfy the principle that being the half of the gate area, the gate ratio becomes to 1:1:2. This is not proper because the metal would enter the runner with the same velocity \( V_{\text{max}} \) which already big enough to cause surface turbulence, if ignoring the effect of the filter. The practical case was making
the runner area 1.8 times of the sprue exit, and the gate area 2 times of the sprue exit, then the ratio became 1:1.8:2. That is, whichever the larger one should be chosen when deciding the runner area from both the sprue exit area and the gate area, when contradiction appears.

4.4.1.5 A filter applied in the gate system

By the time after the metal has fallen down to reach the sprue exit, its velocity usually exceeds the critical value for surface turbulence as judged by We number (eq. 4-5). The gating system should then control the resulting fragmentation of the stream, and gather the stream together again so that it can enter the runner afterwards as a compact spreading front with sufficiently low velocity to avoid further breakup of the front. One way to realize this is setting a two-dimensional filter, a wire mesh between the sprue and the runner at the sprue base as shown in Fig. 4.8. The main purpose of using the filter is to reduce the flow rate of the metal and to retain impurities from entering the casting. Therefore, the proper location of the filter is in the runner, immediately adjacent to the sprue exit where the metal has the maximum velocity. The resistance, caused by surface tension while metal penetrating the pores of the filter and friction between the wires and the metal, provides further contribution by choking the metal at the sprue bottom. It then helps the sprue be fully filled at the earlier stage, therefore reduce air entrapment. The effect of a filter in reducing velocity can be practically measured. An easy way to do this is to measure the total filling times of a casting for with and without a filter cases. Calculating $V_{\text{max}}$ by eq. (4 – 4), and using eq. (4-8) to see the influence of the filter on the friction coefficient $\tilde{\phi}$. The author’s experiments with an aluminum-alloy gravity piston shows that a ceramic wire mesh filter together with mold-wall friction can cause 62% reduction in the average flow rate, i.e., $\tilde{\phi} = 0.62$,
while this loss is only 81% ($\eta = 0.81$) without using the filter, which is close to Jhone Campbell proposal for friction only. It is worthy to mention that the $\eta$ value has close relation with the filter type and location, the structure of a gate system itself. The readers are suggested to find the proper value from the existing gating systems, in order to obtain a guide for the future gating system designing. Whether a filter can trap oxide film or not depends on the mesh size and the material of the filter. Attention needs to be paid that filter itself may also cause inclusion (see Fig. 4.11).

![Fig. 4.11 Inclusion caused by a filter](image)

When using the wire mesh type filter, one thing to be noticed is that the possibility for the metal to bypass the filter. When the filter is bended by metal flow in the way as shown in (a) of Fig. 4.12, some metal will not flow through the filter, but enter the cavity with bypass as shown in (b) of this figure. Trying to hold, as many sides of the filter as possible, or applying some kinds of holding pins are the countermeasures to prevent bypass so that the filter can work effectively.

4.4.1.6 The gate

While designing a gate system, the real parameter that needs controlling is the velocity at which metal enters the cavity. If molten metal splashes or fountains into
the cavity, all the efforts made in gating system design were will be ruined. For the
gate system as shown in Fig. 4.8, the metal velocity is at its highest at the sprue exit,
considerably reduced in the runner because of the two right-angle turns in quick
succession and the resistance of the filter. This velocity is expected to be reduced
further at the gate areas by expanding the 2 gates. It is then vital to try to make the
gate have at least the same sectional area as that of the runner. In the case of a
vertical gate succeeding from a vertical runner, the gate and the runner are partially
filled in the beginning of filling, it is then very important to equalize the gate
thickness with the runner so as to obtain a laminar flow before and after the gate.
Nowadays, it is very convenient to check the maximum velocity in the cavity with
computer simulation. The calculated maximum velocity can then be used to
estimate the Weber number in order to judge whether surface turbulence will occur
in the cavity or not.

The next issue to be considered in making the gate design is the possibility of gate
porosity. Especially when the gate is connected to the casting in the way that a T-junction is created, a hot spot will form at the gate area, if the relative thickness of the gate to the casting is not proper. Traditional modulus calculation suggests making the gate section less than one half or more than double that of the casting to avoid gate porosity\[^{15}\]. This simple rule does not consider the local preheating effect of the metal flowing through the gate, and the preheating effect is continuously in the case of permanent casting.

Besides, this simple rule also does not consider the influences of the metal physical properties. On the other hand, computer simulation does not have these kinds of disadvantage. Chapter 6 shows what to do to avoid porosity at a gate area when making a gating system design. Readers who are interested in this issue are also suggested to read the author’s paper presented in the 8\(^{th}\) Asia Foundry Conference \[^{16}\].

4.4.2 Foundry operation

4.4.2.1 Pouring temperature

Traditionally, as long as cold-shut defect does not form, pouring temperature was always tried being made as low as possible to save energy, to reduce gas absorption, and to avoid over-heating of the mold. However, a lower pouring temperature, under which the viscosity becomes higher, not only prevents gas pores from floating up, but also makes it easier for oxide films to stay with the inserts preset in the cavity, while oxide film is always a good nucleation site for porosity formation as being known. Additionally, for Al-Si alloys, if the pouring temperature is too low, primary Si heterogeneously enucleated during mold filling deteriorating casting mechanical properties for its over size caused by long growing time\[^{12}\].

It is true that metal with a higher pouring temperature uses up more energy and
causes more gas absorption. However, with the introduction of water/air cooling to permanent mold casting, gone are the days in which the lower the better while thinking about pouring temperature. The flowing cooling agent (water or oil) can take away the extracted superheat from the mold very quickly, and makes the mold stay in a much lower temperature than in the case of without mold cooling. There is then no need to worry about the overheating of the mold with superheat. In addition, a higher cooling rate caused by intensive mold cooling allows a higher gas content because the gas-pore evolution during solidification becomes more difficult. Moreover, a higher pouring temperature helps to reduce oxide-films defects around the preset inserts, and restrains primary Si to form in flowing condition. Besides, various impurities, which may have chance to be the nuclei of solidification, are melted under a higher pouring temperature. As a result, bulk solidification will start at a bigger super-cooling and brings about a finer microstructure and hence a superior mechanical properties.

4.4.2.2 Mold cooling

There are 2 concepts concerning cooling rate: one is the average cooling rate, \( \frac{dT}{t_f} \), in which \( T \) is the solidification range of the alloy, and \( t_f \) is the local solidification time; the other is the instant cooling rate, \( \frac{dT}{dt} \) calculated at a specific instant, such as at critical solidification fraction. In order to limit the growth of gas pores and to bring about a finer micro-structure, a bigger average cooling rate is usually preferred. But it is not always good to have a bigger instant cooling rate because it corresponds to a bigger instant solidification velocity, which causes equiaxed growth and hence benefits porosity formation. For example, a last solidification area where porosity often forms usually has a higher solidification velocity. This is the reason why some thermal parameter-based criteria use solidification velocity in its denominator (refer to
Table 3.1. Simulations results showing this will be given in Chapter 5.

Local intensive cooling, such as fountain-type mold cooling, by which the average cooling rate at that area is increased can always help to reduce porosity size when feeding metal is available. But cooling cannot eliminate porosity at a final solidification area, if there is no feeding metal. For this reason, it is very important to make a proper cooling plan for a permanent casting cooled at various mold locations, so that the feeding path is guaranteed during solidification. In other words, the starting and running time of the cooling should be well settled in order to realize a uni-direction solidification towards the riser. An improper cooling plan may chock the feeding path and cause porosity that would not occur if without the cooling.

In summary, a higher pouring temperature and intensive mold cooling is the recent trend of gravity permanent mold casting according to the author’s experiences. Therefore, the mold should be made as thin as possible so that it can stay with a high temperature during mold filling but goes down to a low temperature quickly when solidification starts. Besides, countermeasures to compromise gas absorption; such as continuous degassing in higher humidity seasons have sometimes to be taken into consideration whiling applying a higher pouring temperature. For each specific casting, there actually exist an optimizing pouring temperature and a cooling plan, which usually have to be found out by several trials under a general instructive rule.

4.5 Inserts

General speaking insets help porosity formation. It is then more difficult to get a casting free from porosity when it has inserts. Porosity forms around an insert for
the following reasons. (1) There exists a possibility for oxide films to cling on an insert, which are good site for porosity formation. (2) Inserts behave as barrier to feeding. A proper gating system design, changing the direction of fluid flow by tilting the mold and higher pouring temperature, etc., can help to reduce oxide film adhesion to inserts. But hindrance to feeding is something we have to bare and compromise with proper riser system design.

In addition, inserts change the temperature contours of the fields where they are, and hence, solidification pattern will be deformed. Since basically more heat flows through a more conductive material, inserts of different materials have different influence to porosity formation.

4.5.1 Metal inserts (cast iron and steel)

There are usually two types of metal inserts, with and without an inter-metallic layer formed by bonding process. Oxide films will usually stick on the surface of the former insert after being taken out from the bonding liquid. These oxide films are supposed to be washed away during mold filling. The ways to do this are: keeping the insert in a tilting direction during pouring so that the oxide films are easily washed away. The flow rate should be kept in a certain level, so that it has enough power to flow away the oxide films. It needs to be pointed out that flow rate is decided by the gate system designed and the temperature of the metal. For the insert without inter-metallic layer, the only thing needs attention is to keep the surface of an insert clean and free from moisture.

Even if there is no oxide film around the inserts, it is easier for porosity to form around a metal insert whose conductivity is lower than that of the Al alloy, such as cast iron and steel inserts. This happens because not only the hindering effect of the inserts to feeding, but also due to uneven heat flowing at the insert area (Fig. 4.
13). An example of porosity at the steel and Ni-resistance iron insert is shown in Fig. 4.14.

![Temperature distribution](image1)

**Fig. 4.13** Heat transfer around an iron insert

![Porosity at a metal-insert area](image2)

**Fig. 4.14** Porosity at a metal-insert area

4.5.2 Non-metal inserts (salt, fiber-reinforce material, and sand-core)

There is no difference between a metal and non-metal inserts concerning the barrier effect to feeding, and the role in changing heat flow around the insert since all of them have worse conductivity than Al alloys. But things become simpler for non-metal inserts because there is no oxide film clinging with them before pouring.

Then, attentions to be paid are as followings: (1) keeping the inserts free from
moisture so that no chance for H₂ to form; (2) keeping the surface of the insert as smooth as possible so that it is not easy for oxide films to stick on their surface; (3) creating some permeability to the inserts whenever possible, or making some vents around the inserts, so that the entrapped air caused by meeting of meal fronts around the inserts has chance to escape.

References

(7) A. Pacz, U. S. Patent No. 1387900 (August, 1921).
(11) F. L. Arnold and J. S. Prestley, “Hypereutectic Aluminum-Silicon Alloys


Chapter 5 Preliminary Calculations for Simulation

5.1 The Criterion to Use __ the Niyama Criterion, $G/R^{1/2}$

The theoretical background of computer simulation for foundry process, both mold filling and solidification has been well established. The optimization of gating-system designing, overflow localizing aided by mold filling simulation and solidification-sequence prediction based on solidification simulation, have been widely used. However, there have been no well-accepted ways by which the dispersed porosity of aluminum alloys can be precisely predicted. To predict the locations where porosity would occur, a judging criterion is needed. The popular thermal-parameter based criteria from the literature review covering the period of the year 1996-2000 were summarized in Table 3.1 (Chapter 3). As can be seen from the table that most criteria contain temperature gradient, $G$ in their numerator, and solidification velocity, $V$ in their denominator. All of them can be reduced to the form, $G^n/V^m$ where $n$ varies in the range of 0~2 and $m$ varies in the range of 0.25~1. There has been so far no agreement on which criterion is the best in predicting dispersed porosity of aluminum alloys, as a matter of fact, however, the Niyama criterion, $G/R^{1/2}$, has been widely implemented in current well-used commercial simulation software. The reasons of the popularity of this criterion were discussed in chapter 3 (section 3.2.3.1). Not only because the values of this criterion are easy to be confirmed by experimental measurements, but also because $G/R^{1/2} = (G/V)^{1/2}$, while $G/V$ is the most important parameter governing the constitutional undercooling, so that the range of mushy zone, and hence decides the way of solidification. It is well accepted that columnar or equiaxed growth depends on whether $G/V \leq mc_0(1/k-1)/D$ or not. Therefore, this criterion has essentially a close
relation with the solidification process and porosity formation of an alloy.

The author has, therefore, decided to use this criterion in this research, and actually this criterion has been used in daily solidification simulation in the author’s company whenever a porosity problem occurs.

5.2 Things to Be Noticed While Using the Niyama Criterion

5.2.1 The critical value of the Niyama criterion

For steel casting, a critical value of $1 \text{deg}^{1/2} \text{min}^{1/2} \text{cm}^{-1}$ has been proposed by the submitters. This critical value has been proved being discriminative in judging porosity locations for steel castings$^1$. Different from steel casting, an aluminum-alloy (Al-alloy) usually has a lower temperature gradient $G$ in the whole range of solidification, due to its better heat conductivity. As the result, the values of $G/R^{1/2}$ are lower, and the variation of $G/R^{1/2}$ is smaller for an Al-alloy casting than those for a steel-alloy casting. Since most Al-alloys have conductivity more than two times higher than that of a steel alloy, a value of $0.5 \text{deg}^{1/2} \text{s}^{1/2} \text{cm}^{-1}$ is then assumed as the critical value for Al-alloy castings, that is, the instant temperature gradient $G$ of a solidifying area has to be bigger than $1 \text{deg cm}^{-1}$, if the corresponding instant cooling rate of the area is $4 \text{deg s}^{-1}$. An isolated area with $G/R^{1/2}$ values less than $0.5 \text{deg}^{1/2} \text{s}^{1/2} \text{cm}^{-1}$ will be treated as a potential porosity area.

5.2.2 The moment to calculate the Niyama criterion

Since the purpose is to calculate $G/V$ value at the solidification front via the relation $G/V = [G/R^{1/2}]^2$, it is ideal to calculate $G/R^{1/2}$ at the liquid/solid interface, i.e., $fs = 1$ for any element. However, this is not feasible because the cooling rate is very big at the moment when $fs = 1$ due to the disappearance of latent heat. In numerical
simulation, the cooling rate $R$ of every element is calculated in-between the two time steps, big fluctuations will occur to $R$ if it is calculated at $fs = 1$. Such fluctuations will influence the trustworthiness of the Niyama criterion value.

In order to find the proper solid faction ratio, $fs$, at which the Niyama criterion should be calculated, a 12mm-thick Al-alloy (AC8A) plane-casting model with one-dimensional heat transfer to the mold whose thickness was 10 times to the casting thickness (120mm) was used. The element size was 0.2mm, and the heat resistance between mold and casting was set as 10 (cm$^2$s deg/cal), a value usually used for gravity die-casting in daily solidification simulation.

For comparison, the calculations have also been performed with a 1mm-thick Zn alloy (ZDC2) thin-plane die-casting, for which the mold thickness was 50mm, 50 times to the casting thickness. The element size was 0.025mm, and the heat resistance between mold and casting was set as 4.5 (cm$^2$s deg/cal), a value usually used for die-casting in daily solidification simulation. The physical and thermal properties for AC8A and ZDC2 used in the calculations are listed in Table 5.1. Different superheats were set for ZDC2 and AC8A in order to closely simulate die-casting and gravity casting. The calculations were done by a commercial software, named AdStefan3D, originally developed by Tohoku University.

<table>
<thead>
<tr>
<th>Table 5.1 Physical properties used in the calculation</th>
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<tbody>
<tr>
<td>Density g/cm$^3$</td>
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<tr>
<td>------------------</td>
</tr>
<tr>
<td>Al alloy (AC8A)</td>
</tr>
<tr>
<td>Zn alloy (ZDC2)</td>
</tr>
<tr>
<td>Mold (SKD61)</td>
</tr>
</tbody>
</table>

The Niyama criterion values, $G/R^{1/2}$ and the corresponding temperature gradient, $G$
calculated at different solidification fraction ratio (fs) versus the distance from the mold surface for AC8A are summarized in Fig. 5.1 and Fig 5.2. As can be seen from these two plots, the altering aptitude of the Niyama criterion, G/R\(^{1/2}\) reflects the altering aptitude of the temperature gradient, G when fs = 0.7 ∼ 0.9. But fluctuations appears when fs = 1.0. This fluctuation comes from the fluctuations of the cooling rate R as analyzed above (see Fig. 5.3). Fig. 5.4 shows that when fs is less than 0.7, the instant cooling rate decreases continuously with the distance away from the mold surface. But when fs = 0.7 ∼ 0.9, the instant cooling rate increases with the increase of the distance from the mold surface, and reach its maximum at the final solidification area. This result reflects the well-known phenomenon, the acceleration of solidification\(^2\),\(^3\).

**Fig. 5.1** The influence of calculating instant on the Niyama criterion, G/R\(^{1/2}\) (AC8A)

With ZDC2, the altering aptitude of the Niyama criterion, G/R\(^{1/2}\) reflects the altering aptitude of the temperature gradient, G when fs = 0.7 ∼ 0.85, while fluctuations appear when fs < 0.9 (see Fig. 5.5 & Fig. 5.6). Fig. 5.7 shows that before fs < 0.5,
The Influence of Calculating Instant on G

Fig. 5.2 The influence of calculating instant on the temperature gradient, G (AC8A)

The Influence of Calculating Instant on \( R \)

Fig. 5.3 The influence of calculating instant on the cooling rate - \( R \), R (AC8A)

The Influence of Calculating Instant on \( R \)

Fig. 5.4 The influence of calculating instant on the cooling rate - \( R \), R (AC8A)
the instant cooling rate decreases continuously with the distance away from the mold surface. But when \( f_s = 0.7 \sim 0.85 \), the cooling rate decreases first with the increase of the distance from the mold surface, and rises up towards the final solidification area. Although a fluctuation appears for \( f_s = 0.9 \), the acceleration of solidification phenomenon can be clearly seen after \( f_s \geq 0.7 \). This is in good agreement with the results as calculated with AC8A.

![The Influence of Calculating Instant on \( \frac{G}{R^{1/2}} \)](image)

Fig. 5.5 The influence of calculating instant on the Niyama criterion, \( \frac{G}{R^{1/2}} \) (ZDC2)

![The Influence of Calculating Instant on G](image)

Fig. 5.6 The influence of calculating instant on the temperature gradient, \( G \) (ZDC2)
The Different R Values Calculated at Different fs

Fig. 5.7 The influence of calculating instant on the cooling rate, R (ZDC2)

Based on the results of the above preliminary calculations, it is cleared that the proper moment to calculate the Niyama criterion, G/R$^{1/2}$ for porosity prediction is at the final stage of solidification, i.e., at a higher solidification fraction ratio range. And the proper range for different alloys is somewhat different. For AC8A, it is the span between 80% and 90% ($fs = 0.8$~$0.9$) of an element has solidified; while this span for ZDC2 is between 70~85% ($fs = 0.7$~$0.85$) of an element has solidified.

5.2.3 The cooling rate used for calculating the Niyama criterion

So far the Niyama criterion is calculated at a specified moment (a certain $fs$). That is, both the temperature $G$ and cooling rate $R$ are instant values at a specified moment. There actually exists an average cooling rate by any specified $fs$. Fig. 5.8 shows the difference of the $G/R^{1/2}$ for AC8A as calculated with instant $G$ at $fs = 0.8$, 0.85 and 0.9, respectively, and two kinds of cooling rate:

(a) the instant cooling rates $R_{\text{ins}}$ at $fs = 0.8$, 0.85 and 0.9;

(b) the average cooling rates $R_{\text{ave}}$ calculated with $\overline{\Delta T}/tf$, in which $\overline{\Delta T}$ is the solidification range of the alloy, $tf$ is the local solidification time till the corresponding moment ($fs = 0.8$, 0.85 and 0.9);
Fig. 5.8 The G/R$^{1/2}$ values calculated with different cooling rates at fs = 0.8~0.9(AC8A)

Although the plot altering trends for two types of cooling rates are similar, at the location close to mold surface, the Niyama criterion G/R$_{ins}^{1/2}$ has much higher values than the corresponding G/R$_{ave}^{1/2}$ calculated with average cooling rates. This means intensive cooling does not influences the instant cooling rate, R$_{ins}$ at a higher solidification fraction ratio as much as it does to the corresponding average cooling rate, R$_{ave}$. The two kinds of cooling rate, R$_{ins}$ and R$_{ave}$ are compared with Fig. 5.9. From this figure, obviously it is the instant cooling rate, R$_{ins}$ rather than the average cooling rate, R$_{ave}$ that should be used in the Niyama criterion calculation, so that the acceleration of solidification can be reflected, and hence a last solidification area will always have the lowest Niyama criterion values.

The same calculations have been performed with ZDC2 alloy. Fig. 5.10 shows the results. As can be seen from this figure, the values of G/R$_{ins}^{1/2}$ gradually decrease with the increase of the distance from the mold surface, which is reasonable and easy to understood. On the other hand, the changing aptitude of G/R$_{ave}^{1/2}$ is difficult to explain, as a maximum value appears somewhere between the mold surface and the middle of the model.
Fig. 5.9 The comparison between the instant and average cooling rate calculated at $fs = 0.8 \sim 0.9$(AC8A)

Fig. 5.10 The $G/R^{1/2}$ values calculated with different cooling rates at $fs = 0.7 \sim 0.85$(ZDC2)

The altering trends of the two kinds of cooling rates $R_{\text{ins}}$ and $R_{\text{ave}}$ are similar to that of AC8A (Fig.5.11). The average cooling rates $R_{\text{ave}}$ decreases continuously with the increase of the distance from the mold surface, while the instant cooling rates $R_{\text{ins}}$ reach their maximum values at the final solidification area. Again, it is the instant cooling rate, $R_{\text{ins}}$ that reflects the acceleration of solidification phenomenon.
Fig. 5.11  The comparison between the instant and average cooling rate at 
\[ fs = 0.7 \sim 0.85 \) (ZDC2)

With the condition that to calculate the Niyama criterion with instant \( G \) and \( R \) at a moment when most of an element has solidified (\( fs = 0.8\sim 0.9 \) for AC8A & \( fs = 0.8 \sim 0.85 \) for ZDC2, for instance), an area that solidifies last will usually have the lowest value of the Niyama criterion within the casting. This clarification is very important because it gives the base to applying the Niyama criterion to a non-steel alloy. As reviewed in Chapter 3, the foundation of those opinions against using the Niyama criterion to Al-alloys is that a lower Niyama criterion value may be caused by an intensive cooling, therefore a higher cooling rate, which should be good to prevent porosity from formation because it gives no time for \( H_2 \) to precipitate. Now, it is clear that an intensive cooling will usually cause an increase in the average cooling rate, \( R_{\text{ave}} \), and hence prevent porosity from formation. However, an intensive cooling does not necessary cause an increase to the instant cooling rate \( R_{\text{ins}} \) used in calculating the Niyama criterion. In fact, at a last solidification area, the instant cooling rate \( R_{\text{ins}} \) always goes to the highest, even though cooling is certainly not as intensive as at the mold surface area.

In other words, the instant cooling rate, \( R_{\text{ins}} \) used in calculating the Niyama criterion
is actually a replacement of the instant solidification velocity $V$ at the later stage of the solidification. The value of this criterion reflects the dynamical value of the most important solidification parameter $G/V$ at moment when it is calculated. By calculating this criterion for different elements at the same solidification ratio ($f_s$), we can find out those elements that have a relative lower $G/V$ at the final stage of their own solidification. There might be some other reasons for an element to have a lower $G/V$ at the final stage of its solidification, but being in the last solidification areas is certainly one of the reasons. It has become clear from the above calculations that a higher Niyama criterion value comes from a higher instant temperature gradient $G$ and a lower cooling rate $R_{ins}$. It has no direct relation with the average cooling rate, $R_{ave}$, which is influenced by local cooling, and $R_{ave}$ itself decides the local solidification time, and hence a bigger $R_{ave}$ keeps porosity from formation.

5.2.4 The influences of calculation conditions on the Niyama criterion values

5.2.4.1 Element sizes

The same 12mm AC8A plate-casting model was used. The calculation conditions were made all the same as in section 5.2.2 & 5.2.3, except for two kinds of mesh sizes (0.1mm and 0.2mm). The calculating moment was set at $f_s = 0.85$. The results were shown with Fig. 5.12 ($E = 100$ µm and $E = 200$ µm represent the two element size). As can be seen, element size has little influence on the Niyama criterion, $G/R^{1/2}$ values.

5.2.4.2 The mold/casting interface heat resistance

This time, the calculations were done with 200 µm-element size, same physical
Fig. 5.12 The influence of element sizes on the Niyama criterion, $G/R^{1/2}$ (AC8A)

properties as listed in Table 5.1, but different mold/casting heat resistance, $H_r = 0$ & $H = 10$ (cm$^2$s deg/cal) were tried. The calculated Niyama criterion values are shown with Fig. 5.13. Fluctuations appear when the heat resistance, $H_r$ was set as zero. This fluctuation comes from the changing of the cooling rate and the extremely high temperature gradient at the location close to mold surface (see Fig5.14 & Fig. 5.15).

Fig. 5.13 The influence of heat resistance, $H_r$ on the Niyama criterion, $G/R^{1/2}$ (AC8A)

For ZDC2, with a low initial temperature ($S_h = 1$deg), the Niyama criterion, $G/R^{1/2}$ is
very low at the location close to the mold surface when heat resistance, $H_r = 0$.

Fig. 5.14 The influence of heat resistance, $H_r$ on the temperature gradient, $G$ (AC8A)

Fig. 5.15 The influence of heat resistance, $H_r$ on the cooling rate, $R$ (AC8A)

This abnormal phenomenon was caused by the extremely high cooling rate. A higher initial temperature ($S_h = 33$ deg) gives a higher Niyama criterion value, but fluctuation appears $H_r = 0$. On the other hand, when $H_r$ was set as 4.5 ($cm^2s deg/cal$), both low and high initial temperatures ($S_h = 1$ deg & $S_h = 33$ deg) give reasonable results for the Niyama criterion values (Fig. 5.16).
Fig. 5.16 The influence of heat resistance, $H_r$ and super heat, $S_h$ on the Niyama criterion, $G/R^{1/2}$ (ZDC2)

It can be drawn out from the above calculations that zero mold/casting interface heat resistance should usually not be used, especially when the superheat, $S_h$ is very small. Otherwise an unreasonable distribution of the Niyama criterion, $G/R^{1/2}$, will be obtained. A proper heat resistance, for example, 10 cm$^2$/s deg/cal for AC8A gravity-casting, and 4.5 cm$^2$/s deg/cal for ZDC2 die-casting, will give a reasonable distribution for the Niyama criterion. That is, the Niyama criterion decreases gradually with the distance away from the mold surface.

On the other hand, an extremely high heat resistance, for example, $H_r = 36$ cm$^2$/s deg/cal for ZDC2 thin-die-casting will give an even more ridiculous Niyama criterion distribution for the extremely low cooling rate $R$ resulted. In short, it is not suggested to apply either zero or extremely high heat resistance in solidification simulation, while using the Niyama criterion to predict porosity formation.

5.2.4.3 The Initial metal temperature

To see the effect of super heat on the Niyama criterion value, two calculations with otherwise the same conditions as shown in Table 5.1, except for in one of the calculations, the initial metal temperature was set as 570°C ($S = 4$ deg) comparing...
to 740°C ($S = 174$ deg). The heat resistance $H_r$ was set as 10 (cm$^2$ s deg/cal). As can be seen from Fig. 5.17, the super heat has almost no influence on the Niyama criterion values. Calculations with ZDC2 have actually given the same result (refer to Fig. 5.16).

![The Super Heat $S_h$ on $G/R^{1/2}$ (AC8A)](image)

Fig. 5.17 The influence of super heat, $S_h$ on the Niyama criterion, $G/R^{1/2}$ (AC8A)

5.3 Correlation Between the Potential Porosity Location and the Niyama Criterion Values

This time, a plate with certain thickness but infinite length and width was used. Heat transfer is possible only in its normal direction, i.e., one-dimensional heat-transferring model. The physical and thermal properties of the aluminum alloy, AC8C, as written in Table 5.1 were used. Heat resistance between mold and casting was set as 10 (cm$^2$ s deg/cal). As can be surmised before the calculation, the center-layer of the plate will be the final solidification area, and porosity may appear at this area if no feeding is available.

When the calculating instant was set at $fs = 0.85$, the instant temperature gradient $G$ was the lowest and the instant cooling rate $R$ was the highest at the central-layer area [(b) & (c) in Fig. 5.18]. As the result, the central - layer where $G/R^{1/2} < 0.5$
(deg$^{1/2}$ $s^{1/2}$ $cm^{-1}$) was predicted as the potential porosity area [(a) in Fig. 5.18]. On the other hand, when the calculating instant was set at $fs = 0.25$, the central - layer is not the area where the instant cooling rate is the highest [(d) in Fig. 5.18].

In short, with the condition that to calculate the instant temperature gradient $G$ and cooling rate $R$ at a higher solidification fraction ratio ($fs = 0.85$), a correlation between the potential porosity area and lower Niyama criterion values has been confirmed. At the same time, this calculation has, in some extent, given the foundation of setting 0.5 (deg$^{1/2}$ $s^{1/2}$ $cm^{-1}$) as the critical value of porosity prediction for aluminum permanent mold castings.

(a) $G/R^{1/2}(fs=0.85)$  (b) $G$ (fs=0.85)  (c) $R$ (fs=0.85)  (d) $R$ (fs=0.25)

Fig. 5.18 The calculated $G$, $R$ and $G/R^{1/2}$ of an infinite plate with one - dimensional heat flow (AC8A)

**Reference**


Chapter 6   Reducing Porosity for Aluminum Permanent Mold Castings in Daily Production Aided by Simulation

Although belonging to the same CAE (computer aided engineering) realm, foundry simulation, either fluid-flow or solidification simulation, has not yet become a routine operation in manufacturing industries like FEM for stress analysis and lifetime prediction. It is so not because foundry simulation does not have a good credit as FEM does, but for foundry process can usually be controlled based on accumulated experiences, while stress and lifetime prediction cannot easily be done like that. Nevertheless, whenever a big trouble appears in foundry, abnormal high rejection rate that causes big economical loss, for instance, people will usually turn to simulation for help. This is the reason why a manufacturing industry with a certain scale usually has at least one kind of foundry simulation software.

It is said that simulation is just like a scissor, whether it is helpful or not depends on the person who uses it. This explains why foundry simulation is made good use of in some companies, but not really well used, in other words, having simulation software is just like a decoration in some other companies. As mentioned earlier, porosity problem is a forever topic for Al-alloy permanent-mold castings. Having recognized the virtuality of the Niyama criterion and the proper moment to calculate this criterion for permanent mold casting, porosity prediction with this criterion and designing optimization aided by solidification simulation has started playing its role in daily manufacturing of the author’s company. This chapter presents some typical calculation examples, showing how the porosity problems had been solved in production.

6.1 Porosity Around a Non-aluminum Insert  __ Porosity at the Ring-carrier
Area of a Gravity Al Piston

Enveloped casting technique has long been applied to Al-alloy castings in order to improve local mechanical properties. For high-duty Al-alloy pistons, a high-Ni resistant gray-iron ring-carrier is often preset at the first-groove area (see (a) of Fig. 6.1) after being bonded with an Al-alloy similar to the piston alloy. However, porosity at the ring carrier area has long been a big headache for foundrymen in daily production (see (b) of Fig. 6.1). The genuine reason of this porosity problem has been clarified with the help of solidification simulation.

![Diagram of ring-carrier insert in piston and porosity at the ring-carrier area](image)

(a) a ring-carrier insert in piston  (b) porosity at the ring-carrier area

Fig. 6.1 Porosity at the ring-carrier area of an Al-alloy piston

A piston top partial model with the ring-carrier enveloped cast was used. Same as in Chapter 5, the commercial software, AdStefan3D was used. The heat resistance between mold and casting was set as 10 (cm² deg/cal). The critical value of the Niyama criterion was set as 0.5(deg¹/² s¹/² cm⁻¹). The physical and thermal properties used for the calculations are listed in Table 6.1.
Table 6.1 Physical properties used in the calculation

<table>
<thead>
<tr>
<th>Material</th>
<th>Density g/cm³</th>
<th>Thermal conductivity cal/cm s deg</th>
<th>Heat capacity cal/g deg</th>
<th>Heat of fusion cal/g</th>
<th>Liquidus temp. deg</th>
<th>Solidus temp. deg</th>
<th>Initial temp. deg</th>
</tr>
</thead>
<tbody>
<tr>
<td>Casting (AC8A)</td>
<td>2.72</td>
<td>0.28</td>
<td>0.23</td>
<td>93</td>
<td>566</td>
<td>538</td>
<td>740</td>
</tr>
<tr>
<td>Ring-carrier (Ni-gray iron)</td>
<td>7.3</td>
<td>0.095</td>
<td>0.48</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>550</td>
</tr>
<tr>
<td>Outer mold (SKD61)</td>
<td>7.8</td>
<td>0.102</td>
<td>0.10</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>200</td>
</tr>
<tr>
<td>Top mold (SKD61)</td>
<td>7.8</td>
<td>0.102</td>
<td>0.10</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>250</td>
</tr>
<tr>
<td>Riser bush (Ceramic)</td>
<td>3.26</td>
<td>0.04</td>
<td>0.24</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>250</td>
</tr>
</tbody>
</table>

The distribution of the Niyama criterion calculated at $fs = 0.85$ is shown with Fig. 6.2. As can be seen there is a band around the ring-carrier where the Niyama criterion is below the critical value, $0.5 \text{ (deg}^{1/2} \text{s}^{1/2} \text{cm}^{-1})$ that predicts a potential porosity area.

![Fig. 6.2  The Niyama criterion value at the land area of a piston](image)

Further calculations were performed with different ring-carrier’s initial temperatures, the low-value Niyama criterion band did not change when the initial temperature of
the ring-carrier was changed from 550 to 680 °C. However, this band area disappeared completely when the ring-carrier was removed from the calculating model. These calculation results are in good agreement with the practical situation as observed in daily production.

(1) Changing the bonding liquid temperature of the ring-carrier did not do any help to reduce porosity at the ring-carrier area.

(2) Piston without a ring-carrier has no porosity problem at the corresponding area; and removing the ring-carrier from a piston that has serious porosity problem at the ring-carrier, solved the porosity problem completely (ref. Fig. 6.3).

It is then clear that the porosity at the ring-carrier area of a piston originates from the existence of the ring-carrier. But why the existence of a ring-carrier causes a low-Niyama criterion band, and hence land-porosity problem? Records of the temperature gradient G, cooling rate R, and the Niyama criterion G/R\(^{1/2}\) calculated with the model shown in Fig. 6.2 were examined. It is found that the low-value of G/R\(^{1/2}\) at the ring-carrier area is a result of the low-value of G beside the ring-carrier. Taking an example of two points (a1) and (b1) away from the mold surface with a
same distance, but point (a1) is beside the ring-carrier while point (b1) is away from the ring-carrier (Fig. 6.4).

The calculated temperature gradient G, cooling rate R and G/R at fs = 0.85 for point (a1) and (b1) are compared with Fig. 6.5. As can be seen that G at point (b1) is more than 6 times of G at point (a1), while R at (b1) is only 1/3 of that at (a1). As the result, the Niyama criterion at point (a1) is much lower, less than 1/10 of that at point (b1).

Fig. 6.4 The mesh model without cooling at the ring-carrier area

Fig 6.5 Changing of the Niyama criterion G/R^{1/2} values around a ring-carrier
The reason why a point beside the ring-carrier has a low temperature gradient $G$ is not difficult to understand. From Table 6.1, it can be seen that the heat conductivity of the ring-carrier is only 1/3 of the aluminum casting. During solidification, more heat will pass the aluminum surrounding the ring-carrier as already shown in Fig. 4.13. This is the reason why point (a1) in Fig. 6.4 has a much lower temperature gradient $G$, and hence a low Niyama criterion value, $G/R^{1/2}$ than point (b1).

But it has to be made clear that the surrounding area of the ring-carrier is not a final solidification area. The starting of solidification and the local solidification time of these two points are compared with Fig. 6.6.

![Comparison of the Starting and Local Solidification Time of Point a1 & b1](image)

**Fig. 6.6** The starting and local solidification time of point a1 & b1

As can be seen, point a1 started solidifying earlier and its local solidification time was much shorter than that of point b1. This clarification is important, because it shows that a potential porosity area is not necessary being an area that solidifies last, and what is more important is that the Niyama criterion is discriminative in predicting porosity at such an area.

Having known the genuine reason of porosity at a ring-carrier area, the question comes to how to solve it. It is clear that any measure that increases the temperature gradient $G$ at the later stage of solidification will help to raise $G/R^{1/2}$, and
hence reduce porosity. Another calculation was made with the same 3D model as shown in Fig. 6.4, but insert a cooling pipe in the mold near the ring-carrier area (Fig. 6.7). The resulted Niyama criterion distribution is shown with Fig. 6.8.

As can be seen that the low-value Niyama criterion band has slightly been reduced comparing with Fig. 6.2. The comparison of the temperature $G$, cooling rate $R$ and the Niyama criterion $G/R^{1/2}$ for with and without the local cooling is made in Fig. 6.9. Obviously, a local cooling at the ring-carrier area has much more effect in increasing the temperature gradient $G$ of a point beside the ring-carrier ($a1 & a2$) than it does to a point away from the ring-carrier ($b1 & b2$). But there is only a little change in the
cooling rate. As the result, the Niyama criterion $G/R^{1/2}$ values at the ring-carrier area will be increased with the local cooling.

![The Influence of Cooling on G, R and G/R^{1/2} Values (at $f_s = 0.85$)](image)

**Fig. 6.9** The influence of local cooling on $G$, $R$ and $G/R^{1/2}$ values

Confirming tests based on this calculation have been performed for a piston with high rejection caused by porosity at the ring-carrier area. As expected, the rejection rate has been reduced to 1/3, and the porosity problem has been admirably solved. Worth to mention here that except for the low temperature gradient $G$ in the solidifying front at the ring-carrier area, which is one of the essential reasons of porosity at this area, oxide inclusions brought in by the ring-carrier from the bonding metal plays its part in causing defects at the ring-carrier area. This topic has already been discussed in section 4.5.1.

### 6.2 Porosity at a T-junction Area _ Porosity at the Ingate Area of a Gravity Al Piston

In daily production of the author's company where Al-alloys pistons are cast with permanent molds, porosity at the ingate area whose structure is similar to a T-junction, namely ingate porosity, accounts for more than half of the foundry's
rejection. In order to systematically figure out the countermeasures by which the ingate porosity can be eliminated or driven out from the piston, a series of solidification calculations, examining the influences of the ingate structure and mold temperature, have been planned and performed.

6.2.1 The influences of the T-junction’s structure

Fig. 6.10 shows the section of a T-junction. The thickness of the T-junction’s head is represented with $D_0$, while its leg thickness is represented with $D_1$. The ratio $D_1/D_0$ was defined as thickness ratio. According to the traditional modulus calculation, there will be no porosity problem at the T-junction area, if $D_1/D_0 \leq 0.5$ (the leg having the cooling fin effect), or $D_1/D_0 \geq 2$ (the leg having the feeding effect). But troublesome porosity problem will certainly occur if $D_1/D_0 = 1$. In other words, when the T-junction shaped structure is applied to an ingate area, the ingate thickness should be designed either less than half or bigger than twice of the product thickness to avoid ingate porosity. However, this requirement is not easy to satisfy, as the ingate will be either too thin that creates jet-flow in mold filling, or too thick that causes difficulty to knock off, if designed according to this rule. In fact, this simple rule needs modification while being applied to ingate area because the locally repeatedly heating of the ingate area caused by metal flows was not considered in the traditional modulus calculation. And the modulus method only considers heat transferring but leaves the influence of the material properties untouched. Simulation results can reflect both the preheating effect of melt flow from mold filling and the influence of the material properties. For example, instead of 0.5 for
all materials, the critical thickness ratios of a T-junction free from porosity calculated by simulation for casting iron and the aluminum alloy AC4C are different, \( D_1/D_0 = 0.78 \) for AC4C, but \( D_1/D_0 = 0.45 \) for cast iron\(^{11,21} \). It has also been reported by the same author that the cooling-fin effect of the thin “leg” to a T-junction is more obvious with AC4C than with cast iron.

A series of calculations, with a T-junction model long enough to avoid end effect and similar to the ingate structure of a piston in intersecting angle and radius as shown in Fig. 6.11 was used. \( D_0 \) was fixed at 1cm, while \( D_1 \) changes from 0.6cm to 1.6cm gradually. The physical and thermal properties used for the casting model and the mold were the same as shown in Table 5.1. The distributions of the Niyama criterion calculated at \( f_s = 0.85 \) for different \( D_1/D_0 \) were shown with Fig. 6.12. An area whose Niyama criterion is lower than the critical value, 0.5 (deg\(^{1/2}\) s\(^{1/2}\) cm\(^{-1}\)) and isolated at the T-junction area is called a hot spot where porosity will occur. From Fig. 6.12, the followings can be summarized:

(1) When \( D_1 < 0.7D_0 \), there is no hot spot problem with the T-junction.

(2) When \( D_1 = D_0 \), a hot spot forms at the T-junction area.

(3) With the increase of \( D_1 \), the hot spot gradually move towards the leg side.

(4) When \( D_1 = 1.2D_0 \), the hot spot starts to appear at the leg side.

(5) When \( D_1 = 1.4D_0 \), the hot spot starts to connect with the last solidification area of the leg so that the leg can act as a feeder in solidification, leaving the head free from porosity.
Fig. 6.12 The transition of hot spot location with the change of the thickness ratio, $D_1/D_0$.

This simulation shows that ingate porosity of an aluminum piston cast with permanent mold can be avoided by making the ingate at least 1.2 times thicker than the skirt. However, in the practice situation, the ingate of the piston is connected not only to the skirt part, but also to the thick head part as shown in Fig. 6.13. Then, there is a turning point regarding to the thickness ratio $D_1 / D_0$, i.e., $D_1 / D_0 > 1$ for the lower part of the skirt but $D_1 / D_0 < 1$ for the upper part of the skirt. Solidification simulation with this model shows that there will be no porosity problem on the lower part of the skirt surface when $D_1 / D_0 = 1.2$, but an area, whose $G/R^{1/2}$ values are...
lower than 0.5 (deg$^{1/2}$ min$^{1/2}$ cm$^{-1}$), crosses the finished machined surface, which indicates the potential ingate porosity location on the finished machined surface (Fig. 6.14).

This predicted location is in excellent agreement with the ingate location of a piston with similar profile. The result of this calculation has shown us that the ingate porosity cannot be totally avoided even if the ingate being designed with the proper thickness ratio as instructed by the T-junction model simulation. Then, what can be done to solve the ingate porosity problem?

6.2.2 The Influence of the mold temperatures

Since ingate porosity is simply a heat-transferring problem, it is hopeful to solve it by controlling the mold temperature at either the ingate or the skirt side. Several calculations were performed to see how the porosity location changes with the changes of mold temperatures on both ingate and skirt side. It has been found out that raising the mold temperature at the ingate side can help move the porosity out
of the piston. The calculation was then done by locally adding insulating material to the ingate side. The location of ingate porosity predicted is shown with Fig. 6.15.

![Finished machined surface](image)

Fig. 6.15 The effect of insulating ingate side on ingate porosity

Comparing with Fig. 6.14, the porosity area where the Niyama criterion is lower than the critical value moved towards the ingate side, which will be cut off together with the ingate system afterwards. In practical case of piston production, the mold temperature at the skirt side is actually higher than that at the ingate side, which makes it even more necessary to raise the mold temperature at the ingate side by locally insulating.

Based on this calculation, the mold of one piston that has higher ingate rejection rate was modified by locally adding insulating material at the ingate side. Such a modification has reduced the rejection rate of ingate porosity to 1/4. Fig. 6.16 shows the sectional pictures before and after the modification. An effective way of reducing porosity at the ingate area whose structure is similar to a T-junction has then been firmly established.
6.3 Centerline Porosity and Porosity at a Hot Spot Area _ Porosity at a Final Solidification Area of an Al Squeeze Casting

It is the excellent reliability in predicting the centerline porosity of steel casting that has made the Niyama criterion well known. In literature, experiments with the typical aluminum alloy, A356 have proved that the minimum density (or maximum porosity) was located in the same position that exhibited the minimum value of the Niyama criterion (ref to section 3.2.3.2). From the calculations performed in Chapter 5, it is clear that a final solidification area of an Al-alloy casting has usually a lower Niyama criterion value. Therefore, there is no doubt about the discriminability of the Niyama criterion to Al-alloy castings in predicting porosity at a hot spot or along a centerline, which solidify last. Moreover, in addition to prediction, the real meaning of simulation is to give suggestions with which a hot spot can be moved or removed, and dispersed porosity along a centerline can be reduced.

The effects of changing the casting designing, such as modifying the structure and/or location of the gating system, changing the volume and/or location of the riser
system, etc., can of course directly be observed with simulation calculations. For some specific products to which no many casting designs are available due to the specific limitations, adjusting the mold cooling can be an effective way to solve porosity problem at either a hot-spot area or along a centerline. One example is given in below. Porosity inside a cylinder-shaped casting of AC8A made by squeeze has once caused big rejection due to the failure in the leakage test (Fig. 6.17).

![A squeezed cylinder with leakage problem caused by porosity](image)

Solidification simulation has been performed with mold filling based temperature field. The calculation conditions were all the same as used in section 6.1 ($H_r = 10 \text{ cm}^2\text{s deg/cal}$ and Table 6.1 for physical properties). The predict porosity location (hot spot) was in excellent agreement with the practical porosity location, when the Niyama criterion was calculated at $f_s = 0.85$ (Fig. 6.18). However, when the Niyama criterion was calculated at $f_s = 0.25$ (Fig. 6.19). The right side of the casting was also predicted as the potential porosity area, which is not in agreement with the practical situation (comparing Fig. 6.19 with Fig. 6.17). Porosity along the centerline of the horizontal gate was predicted for both calculating instants ($f_s = 0.85$ & $f_s = 0.25$), but this problem is out of the consideration, as the gate will after all be removed.
Further calculations were made by reinforcing the mold cooling inside the core at the gate side. The predicted hot spot has parentally been reduced as shown in Fig. 6.20. The result of the simulation was applied to the practical production in the way that adding one more fountain-type cooling inside the core at the gate side. The porosity area has been reduced an acceptable level (see Fig. 6.21). And the leakage problem caused by the inter-connecting porosity at the gate side has been solved. The predicted porosity area was in excellent agreement with the experimental result (Fig. 6.20 & Fig. 6.21).
Fig. 6.20 The predicted porosity after reinforcing mold cooling inside the core

Fig. 6.21 The effect of reinforcing mold-cooling porosity at a hot spot

Reference


Chapter 7 Conclusions

Regarding to the porosity defects of Al-Si alloy castings made with permanent mold, three aspects have been clarified with this research.

1. The porosity defects of Al-Si alloy castings made with permanent mold can be predicted.

2. In order to reduce porosity defects, every step of the casting making, including mold design, melting process and the foundry process need to be strictly controlled.

3. Computer simulation with a solidification software, to which the Niyama criterion, $G/R^{1/2}$ is integrated, is a useful tool in predicting and reducing porosity.

7.1 Porosity Prediction

(1) There have formed three approaches: modulus and equisolidification method, criterion function method, and direct numerical simulation method, with which shrinkage and porosity can be predicted.

(2) Modulus method can be used as a quick guide in the initial stage of mold designing.

(3) Equisolidification method is reliable in predicting gross shrinkage and porosity at a final solidification area.

(4) Direct numerical simulation gives insight into the formation of dispersed porosity among the three approaches. But its application is limited for its complexity and the unknown calculation conditions.

(5) Among the thermal-parameter based criterion, the Niyama criterion has the most popularity for its well-accepted discriminability in predicting shrinkage and porosity of casting steel. And it is easy to verify this criterion with
temperature measurements.

(6) The significance of the Niyama criterion value in solidification comes from the relation, \((G/R^{1/2})^2 = G/V_s\), since the right side of this equation is a very important solidification parameter, if not the most important.

7.2 The Influences of Controlling Parameters in Foundry

(7) The minute amount elements, such as Ti, Zr, and V, which have the effect of refining the primary alpha-dendrite, help porosity formation, if their sum total is over a 0.30%.

(8) P-inoculation to refine primary Si also has effect in reducing porosity. However, if the amount of P added is too much, over 400ppm for instance, a serious contagion, which ruins the outer-looking of the machined product, will occur.

(9) A permanent mold with a bad gating and riser system will cause congenital porosity that is difficult to be removed.

(10) Higher pouring temperature and a well-planned intensive cooling is, in general, beneficial to reducing porosity.

(11) It is easier for porosity to form around a non-Al insert preset in the cavity before pouring.

7.3 Reduce Porosity of Al-alloy Permanent Casting Aided by Computer Simulation

(12) It is important to define the instant at which the Niyama criterion is calculated to the moment while an element is at the location beside the liquid/solid interface, since the virtuality is to calculate \(G/V_s\) via the relation
$G/V = (G/R^{1/2})^2$. Considering the accuracy of the numerical calculation, it is suggested to calculate the Niyama criterion at $f_s = 0.80\sim 0.90$ for Al-alloys, and $0.70\sim 0.85$ for Zn alloys.

(13) When calculating the Niyama criterion, rather than the average temperature gradient $G_{\text{ave}}$ and cooling rate $R_{\text{ave}}$ covering the solidification range, the instant temperature gradient $G_{\text{ins}}$ and cooling rate $R_{\text{ins}}$ should be used.

(14) A final solidification area, such as a hot spot or a centerline area, usually has a low Niyama criterion value.

(15) The area around a non-Al insert usually has a low Niyama criterion value.

(16) An effective way of reducing porosity problem is to reinforce local cooling, by which the Niyama criterion value can be usually increased though the increase of both $G_{\text{ins}}$ and $R_{\text{ins}}$.

(17) While making a designing change or/and making a cooling-plan, examining the transformation of the Niyama criterion, $G/R^{1/2}$ by computer simulation is an effective way to avoid or reduce porosity.
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## List of Publications

**Papers**

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<th>Journals</th>
<th>Authors</th>
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## Discourse

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<td>1</td>
<td>凝固シミュレ- ションによるアルミニウム (AC8A) ピストン「塩先巻」の低減</td>
<td>日本鋳造工学会第141回日本全国講演大会 No. 50</td>
<td>2002年10月</td>
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