On the probabilistic nature of high-pressure die casting

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ABSTRACT

This article unmasks the probabilistic nature of high-pressure die casting; specifically, the cause of scatter in the tensile ductility of die-cast Al8Si0.4Mn0.3Mg (wt.%) alloy. Scatter in tensile ductility is related to the size of large pores and non-metallic inclusions. We propose that these non-metallic inclusions form during the pyrolysis of commercial plunger lubricants, and that these large pores derive from dilatational strains introduced during semi-solid deformation. The apparent randomness of pore formation is thus ascribed to the heterogeneous nature of the semi-solid network. Reducing heat loss in the shot chamber is shown to promote a more homogeneous grain structure, leading to a decrease in the maximum pore size from 1.32 mm to 0.37 mm, and an increase in the minimum tensile ductility from 6.8% to 9.4%.

1. Introduction

High-pressure die casting (HPDC) is widely used in the manufacture of light alloy structures intended for use in automobiles and aeroplanes, due in part to its high dimensional accuracy and suitability for robotic automation. HPDC is characterised by the high-speed injection of molten metal into a sealed mould cavity, followed by solidification due in part to its high dimensional accuracy and suitability for robotic automation. HPDC is characterised by the high-speed injection of molten metal into a sealed mould cavity, followed by solidification under an applied pressure of 30 ~ 100 MPa [1]. The process thus elicits an array of rich and complex phenomena including fluid turbulence, solidification, and semi-solid deformation. Combined, these phenomena make HPDC somewhat probabilistic in nature: the tensile ductility of die-castings is notoriously inconsistent [2–4]. This inconsistency leads to high scrap rates, and increased safety factors for component design. Although the tensile ductility of die-castings has been linked to various casting defects—porosity [5–8], oxides [8,9], and sludge intermetallic particles [9–11] to name a few—the underlying cause of variability remains enigmatic.

Previous investigations into the reliability of die-castings have generally pursued one of three lines of inquiry. The first derives empirical correlations between the tensile ductility and various measures of microstructural uniformity (e.g. bulk porosity content [3,4,12,13], maximum pore size [5–7], grain size [13–15], eutectic fraction [3,4,10]). The second relates the scatter in tensile ductility to statistical variations in melt quality (e.g. chemical composition [11,16], gas content [8], number of inclusions in the melt [9]). The third considers the stochastic nature of fluid flow, and the subsequent encapsulation of air and oxides [17–20]. Although porosity is a recurrent theme in these studies, its formation mechanism is widely disputed [4,9,13]. Tian et al. [9] relate the amount of porosity to the number of inclusions in the melt. Li et al. [13] observe a linear correlation between the volume fraction of porosity and the fraction of primary α-Al, grains solidified in the shot chamber. Dong et al. [4] found that the use of vacuum can lead to fewer pores and improved tensile properties. However, these studies assume that porosity is the main source of scatter in tensile ductility. In light of such findings, it would be interesting to determine whether or not this assumption is valid.

Previously [21,22] we simulated the HPDC process using the finite element method (FEM) under the ProCAST (ESI Group) software platform; an optimum plunger speed profile was derived in Ref. [22] based on predictions from the model. Here, we perform HPDC experiments under both Baseline and Optimised conditions to identify the primary source of scatter in tensile ductility. Based on our findings, we explain the seemingly random nature of pore formation in the HPDC process.

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2. Experimental

We begin by defining a set of casting parameters and operating conditions that are representative of commercial foundry practice (see Appendix). This article does not concern the state of the art, but rather aims to reproduce the variability encountered in industry. The established process was used to produce roughly 100 tensile specimens; the mechanical properties of these specimens will serve as a baseline for later benchmarking. A similar number of specimens were also produced under Optimized conditions, as will be discussed later.

2.1. Casting procedure

Al8Si0.4Mn0.3Mg (wt.%) alloy (40 kg per crucible) was melted in an electric resistance furnace and held at 750 °C for 30 min to maintain a uniform composition distribution. The melt was degassed using a conventional rotary degassing unit for 10 min with a stirring speed of 350 rpm and an argon flow rate of 4 l/min. Although other, more effective, techniques are available (e.g. intensive shearing [8] and ultrasonic processing [23]) rotary degassing was selected as it is most representative of commercial foundry practice. After degassing, the liquid was manually ladled into the shot chamber of a FRECH 4500 kN locking force cold chamber HPDC machine to produce ASTM standard E8/E8M tensile specimens [24]. The die geometry used in this work is shown in Fig. 1. The pouring temperature was maintained at 680 ± 5 °C using a thermocouple. The temperature of the die and shot chamber were held at roughly 200 °C and 180 °C, respectively.

Fig. 2 shows the plunger speed profiles considered in this work. Notable differences lie in the speeds attained at displacements of 50 mm and 375 mm (Baseline {0.2 ms⁻¹, 0.3 ms⁻¹} and Optimized {0.4 ms⁻¹, 0.6 ms⁻¹, respectively). The Baseline profile was defined according to recommendations from the supplier of the HPDC machine (FRECH). Previously [21,22], we simulated the HPDC process using a FEM model developed under the ProCAST software platform. An optimal plunger speed profile (Fig. 2) was derived in Ref. [22] based on the amount and distribution of oxides, entrapped air, and shrinkage porosity predicted by the model. A fast shot speed of 3.6 ms⁻¹ was used for both plunger speed profiles based on preliminary trials. A pressure of 32 MPa was selected for the intensification stage, corresponding to the maximum attainable pressure of the hydraulic actuator.

2.3. Tensile testing and SEM fractography

Tensile tests were performed at room temperature using an Instron 5500 universal electromechanical testing system equipped with a 50 kN load cell, in accordance with ASTM standard E8/E8M [24]. A schematic of the tensile specimen is shown in Fig. 1(c). Tensile data was acquired using a 50 mm extensometer with a ramp rate of 1 mm/min. Roughly 100 tensile specimens were tested for each plunger speed profile. After fracture, specimens representing the distribution of tensile ductility (with particular emphasis on the lower tail) were isolated and the fracture surfaces examined using a LEO 1450VP SEM (Carl Zeiss AG) equipped with energy dispersive X-ray spectroscopy (EDX).

2.4. X-ray tomography

X-ray tomography was performed using an Xradia 410 Versa (Carl Zeiss AG) microtomography system operated with an accelerating voltage of 80 kV and a power of 10 W. An isotropic voxel size was defined with a characteristic length of 3.4 μm. The VGSTUDIO MAX (Volume Graphics GmbH) software was used to perform quantitative analyses on the three-dimensional volumes. An iterative surface determination was performed to isolate the material volume while compensating for local fluctuations in grey value. The VGDEFX module (Volume Graphics GmbH) was used to characterise pores contained in the material, with a minimum pore volume of 8 voxels used in the analysis. Pore size was evaluated using the maximum Feret diameter method (the distance between two parallel tangential planes enclosing the largest dimension of the particle or void). Future reference to particle size will also imply use of the maximum Feret diameter method, with the diameter denoted by D. Preliminary trials revealed that the effects of damage accumulation on pore size and morphology may be ignored due to the brittle nature of fracture. As the average pore sizes (∼50 μm) considered in this work are much larger than the voxel size of 3.4 μm, we may safely ignore the effect of image resolution on pore size measurements.

2.5. Electron backscatter diffraction mapping

Samples for electron backscatter diffraction (EBSD) mapping were prepared to a 0.04 μm finish using standard metallurgical techniques. The samples were then polished for 40 min using a vibrational plate, operated at a frequency of 90 Hz. EBSD was carried out on a Crossbeam 350 FIB-SEM (Carl Zeiss AG) equipped with an EDAX EBSD analyser.

Fig. 1. Die geometry used to produce ASTM standard E8/E8M tensile specimens [24]. (a) Front view. (b) Side view of runner-orifice system. (c) Schematic of tensile specimens (all dimensions in mm).
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3. Results

3.1. Mechanical properties

Fig. 3 shows the tensile properties of specimens produced under Baseline and Optimized conditions; 95% confidence ellipses are shown for two-dimensional normally distributed data. Average values and standard deviations are provided in Table 1. The average values and standard deviations are unchanged by the change in plunger speed profile. However, Fig. 3 shows that the variability in tensile strength and tensile ductility is greatly reduced under Optimized conditions (that is, if we define variability as significant negative deviations from the average property). For example, minimum values of 6.8% and 9.4% are obtained for the ductility of specimens produced under Baseline and Optimized conditions, respectively. A similar trend is also observed for tensile strength, with minimum values of 275 MPa and 282 MPa reported for the Baseline and Optimized conditions, respectively. The contrary is observed for 0.2% proof strength, with minimum values of 143 MPa and 136 MPa obtained for the Baseline and Optimized conditions, respectively.

3.2. Porosity

From each group, the tensile specimen with the lowest ductility was isolated for further examination; X-ray tomography was used to characterise pores contained in these two specimens. In Fig. 4, the sphericity of each pore is plotted against its maximum Feret diameter (D). Here, sphericity refers to the ratio $A_s/A_d$ where $A_s$ is the surface area of the defect, and $A_d$ is the surface area of a sphere with equivalent volume. Values of sphericity lie between zero and unity, with small values indicative of irregular morphology. Average values for pore size and sphericity are provided in Table 2. The average values and standard deviations are unchanged by the change in plunger kinematics. However, the maximum pore size decreases from 1.32 mm to 0.37 mm following the change from Baseline to Optimized. The irregular morphology (low sphericity) of these pores suggests that they do not originate from the expansion of gaseous phases [3–8].

3.3. Inclusions

Fig. 5(a) shows the largest pore (D=1.32 mm) observed on the fracture surface of the lowest ductility specimen in the Baseline group. Large non-metallic inclusions can be seen within the pore; oxides are also observed in the vicinity of these inclusions. Pores are known to

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Table 1

| Condition   | 0.2% Proof strength (MPa) | Tensile strength (MPa) | Tensile ductility (%) |
|-------------|----------------------------|------------------------|-----------------------|
| Baseline    | 148 ± 3 (min. 143)         | 292 ± 4 (min. 275)     | 11.6 ± 1.5 (min. 6.8) |
| Optimized   | 144 ± 3 (min. 136)         | 294 ± 4 (min. 282)     | 11.9 ± 1.4 (min. 9.4) |

Table 2

| Condition     | Avg. Pore size (mm) | Avg. Sphericity |
|---------------|---------------------|-----------------|
| Baseline      | 0.05 ± 0.02 (max. 1.32) | 0.61 ± 0.07 (min. 0.12) |
| Optimized     | 0.05 ± 0.01 (max. 0.37) | 0.59 ± 0.07 (min. 0.29) |

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Fig. 3. Tensile properties of specimens produced under Baseline and Optimized conditions. 95% confidence ellipses are shown for two-dimensional normally distributed data (CI: 95%). (a) 0.2% proof stress vs. tensile ductility. (b) Tensile strength vs. tensile ductility.
nucleate on non-wetted interfaces [25]. Favourable nuclei include the non-wetted surfaces of oxide bifilms and non-metallic inclusions, particularly those comprising low-surface-tension liquids and non-wetted solids [25]. These inclusions, or the accompanying oxides, may act as nuclei for the pores observed in these specimens. Though inclusions in the lowest ductility specimens are relatively small, large non-metallic inclusions (0.4 \(\sim\) 0.7 mm) were observed on the fracture surface of other specimens with a tensile ductility in the range of 8 \(\sim\) 12%. A representative micrograph of such an inclusion is shown in Fig. 5 (b).

### 3.4. Grain structure

Fig. 6 (a and b) shows inverse pole figure (IPF) maps for the Baseline and Optimized samples, produced via EBSD mapping. The grain structure is shown to comprise a mixture of large dendritic primary \(\alpha\)-\(\text{Al}_1\) grains (30 \(\sim\) 150 \(\mu\)m) solidified in the shot chamber, and smaller globular-rosette primary \(\alpha\)-\(\text{Al}_2\) grains (\(\sim\) 10 \(\mu\)m) formed in the die cavity. Such differences in size and morphology are due to the lower cooling rates (\(\sim\) 10 Ks \(^{-1}\) [14]) in the shot chamber compared to those in the die cavity (500 \(\sim\) 1000 Ks \(^{-1}\) [14]). Additional information regarding the nature of primary \(\alpha\)-\(\text{Al}_1\) grains and primary \(\alpha\)-\(\text{Al}_2\) grains can be found in Refs. [14, 27]. Increasing the plunger speed from Baseline to Optimized induces a refinement of both large primary \(\alpha\)-\(\text{Al}_1\) grains and smaller primary \(\alpha\)-\(\text{Al}_2\) grains, which can be ascribed to reduced heat loss in the shot chamber. Fig. 6 (c and d) show grain size distributions corresponding to the IPF maps in Fig. 6 (a and b). Grain size distributions are often observed to approximate a lognormal shape, a phenomenon that has previously been attributed to the time-dependent kinetics of crystal growth processes [26]. Accordingly, we present grain size in terms of its
natural logarithm (i.e. ln D). The two distributions are multimodal—an observation that is more pronounced in the Baseline distribution than in the Optimized distribution. In accordance with previous studies [27,28], a threshold size of 30 μm was used to distinguish primary α-Al1 grains from primary α-Al2 grains. Primary α-Al1 grains are, on average, smaller in the Optimized condition (42.0 ± 13.8 μm) than in the Baseline condition (47.5 ± 14.0 μm). A similar trend is observed for primary α-Al2 grains, with average values of 9.0 ± 5.4 μm and 8.0 ± 5.0 μm reported for the Baseline and Optimized conditions, respectively. Additionally, we can infer from the shape of the distributions that a more uniform grain structure is produced under Optimized conditions.

4. Discussion

4.1. Statistical analysis of tensile ductility

Here, the generalized extreme value (GEV) distribution is used to model the limiting distribution of tensile ductility for specimens produced under Baseline and Optimized conditions. A detailed description of the GEV distribution is provided in Ref. [29]. The cumulative distribution function of the GEV distribution is

\[ G(x) = \exp \left( - \left[ \frac{1 + \xi (x - \mu)}{\sigma} \right]^{1/\xi} \right), \quad x > \mu \]

where \( \mu, \sigma, \) and \( \xi \) describe the location, scale, and shape of the distribution, respectively. \( G(x) \) is equivalent to the Gumbel, Fréchet, and Weibull functions when \( \xi = 0, \xi > 0, \) and \( \xi < 0, \) respectively. The quantile function is obtained by inverting (1) for \( G(X) = 1 - p: \)

\[ X = \mu - \frac{\sigma}{\xi} \left[ -\log \left( 1 - p \right) \right]^{-1/\xi}, \quad \text{for } \xi \neq 0, \]

\[ X = \mu - \sigma \log \left( -\log \left( 1 - p \right) \right), \quad \text{for } \xi = 0 \]

where \( X \) is the return level associated with a return period \( m \) and probability \( p = 1/m. \) It is important to note that equations (1) and (2) are defined for the case of maxima, not minima. To model minima, we use the substitution \( x_i = -y_i, \) where \( y_i \) denote our values of tensile ductility (small values of \( y_i \) now correspond to large values of \( x_i \)). Parameter estimates were obtained by maximum-likelihood estimation, with the sign correction \( \tilde{\mu} = -\mu. \) Parameter estimates and 95% confidence intervals are provided in Table 3. The change in plunger kinematics has little effect on the threshold value \( \tilde{\mu}. \) However, the tail behaviour of the two distributions is markedly different. For example, the negative shape parameter \( \xi = -0.481 \) implies that the Optimized distribution possesses a finite upper bound of \( \sim 9.6\%. \) Conversely, the positive shape parameter \( \xi = 0.102 \) suggests that the Baseline distribution is boundless; the return level plots in Fig. 7 exemplify this disparity. To demonstrate the practical implications of \( \xi, \) let us consider a hypothetical design criterion requiring a minimum ductility of 10%. From equation (2), we obtain scrap-rates of 12.5% and 5.7% for specimens produced under Baseline and Optimized conditions, respectively (i.e. 100/m, for \( X = -10\%). Thus, by changing the kinematics of the plunger we have effectively halved the scrap rates in our imaginary foundry.

Table 3

| \( \xi \)-mean | \( \xi \)-95% Confidence intervals | \( \sigma \) | \( \tilde{\mu} \) |
|----------------|---------------------------------|---------|---------|
| Baseline       | 0.102                           | [-0.070; 0.275] | 1.06    | 12.3    |
| Optimized      | -0.481                          | [-0.589; -0.373] | 1.26    | 12.2    |

4.2. Source of variability in tensile ductility

Previous studies [5–7] have shown that the tensile fracture properties of aluminium alloy die-castings depend on the maximum defect size, with other factors (such as defect morphology, size distribution, and volume fraction) of secondary importance. Fig. 8 plots the tensile ductility of a specimen against the size of the largest defect observed on its fracture surface; defects in Fig. 8 are classified as either pores or inclusions. A power-law relationship of the form \( y = ax^\beta, \) was fitted to the overall data, as shown by the dashpot line in Fig. 8. The power-law accurately captures the behaviour of the empirical data (coefficient of determination \( R^2 = 0.858) \) when the constants \( a \) and \( b \) take values of 4.461 and -0.525, respectively. From Fig. 8, it seems that inclusions \( \sim 0.75 \) mm in diameter have a similar effect on the tensile ductility as pores in the range of 0.90 - 1.50 mm. However, this is likely due to the location of these defects with respect to the specimen free surface, as opposed to the potency of the defects themselves—while pores typically congregate in the central core, non-metallic inclusions are often observed near to the specimen free surface. Strain localises more rapidly at surface defects and sub-surface defects compared to internal defects of an equivalent size [30]. Few oxides were observed on the fracture surfaces of the tensile specimens, and when they were observed they were negligibly small in size; no other abnormalities were observed on the fracture surface of the examined specimens. Based on these considerations, we infer that the observed scatter in tensile ductility is caused by variations in the size of large pores and non-metallic inclusions, the origins of which we now discuss.

4.3. Origin of non-metallic inclusions

Oil-based lubricants are often applied to the plunger tip to prevent wear and seizure. The tribological performance of such lubricants rely on the formation of a low-viscosity film between the contacting surfaces. When liquid metal enters the shot chamber, it reacts with organic compounds contained in the plunger lubricant to produce gaseous phases [31] and a carbonaceous residue [32]—this pyrolysis presents itself as a flame emerging from the pouring hole. This residue may be entrained into the liquid during solidification processing, materializing as non-metallic inclusions in the residual microstructure. Variations in inclusion size may then depend on the rate of build-up, and the viscous forces leading to detachment and breakage.

EDX analysis (see insert of Fig. 5(b)) reveals that these inclusions typically contain C, Na, S, Cl, K, and Ca. Due to its low atomic mass, the presence of C does not substantiate the pyrolysis hypothesis; C is also a common contaminant in electron microscopes. Let us, therefore, turn our attention to the heavier elements: Na, S, Cl, K and Ca. Petrochemical or synthetic based additives are often added to base stocks to impart additional properties to the lubricant. Extreme pressure additives enhance lubricity in high-pressure environments, and often contain compounds of S or Cl [33]. Other additives, such as surfactants and thickeners, are also used in commercial lubricants, often containing compounds of Na, K and Ca [34]. The presence of Na, S, Cl, K, and Ca in the observed inclusions is therefore consistent with the chemical constituents of commercial plunger lubricants. To test the pyrolysis hypothesis, filling tests were performed by allowing the melt to solidify in the shot chamber with the plunger at rest (see Appendix). Large films, that closely resemble the non-metallic inclusions shown in Fig. 5(b), were identified on the surface of the solidified billets (Fig. A1). Hence, we propose that these non-metallic inclusions form during the pyrolysis of commercial plunger lubricants in the shot chamber. One might argue that the aforementioned elements are also found in commercial foundry fluxes; however, no flux was used in our experiments.
4.4. Origin of porosity

Previously we found that a change in plunger kinematics may lead to a decrease in the maximum pore size; however, the average pore size remained unchanged. We also related the observed scatter in tensile ductility to variations in the size of these large pores. Fewer oxides accumulate in the casting under Optimized conditions, due to a calmer free surface during fluid flow [22]. However, no relationship was observed between the tensile ductility of a specimen and its position in the die. Furthermore, the average gas content is lower in the Optimized condition than in the Baseline condition, due to reduced air entrapment in the shot chamber [22]. Although an increase in gas content would certainly expedite pore growth, it is unlikely that a small difference would have a significant effect. Moreover, one would expect the average pore size to also increase following an increase in the average gas content.

An alternative explanation lies in the response of the semi-solid alloy to shear deformation. During the intensification stage, a pressure of 30 ~ 100 MPa [1] is applied to the solidifying alloy to compress gaseous phases and to assist in the feeding of shrinkage strains. Previous studies [35–38] have shown that metallic alloys at 60 ~ 95% solid deform as near-cohesionless granular materials, exhibiting phenomena such as Reynold’s dilatancy and granular rearrangement. Under compression, a pore will either contract or dilate depending on the solid fraction at which deformation occurs [35,36]. At low solid fractions, grains are able to move towards one another, applying an external pressure on the pore, driving closure [36]. At high solid fractions, grains are so densely packed that grain rearrangement leads to shear-induced dilation and pore growth [36]. Kareh et al. [35] report a volumetric increase of ~ 622% for a pore in an Al–Cu alloy deformed at 93% solid. The maximum volumetric strain increases with the solid fraction at which deformation occurs [35,36]. Furthermore, the melt temperature at the end of die filling is 30 °C higher in the Optimized condition compared to the Baseline condition [22]. Thus, the change in plunger kinematics from Baseline to Optimized will lead to a decrease in the maximum volumetric strain encountered during semi-solid deformation. Dilatational strains are highly localised and time variant, and may lead to the formation of a new pore, or the sudden expansion of an existing pore [35,36]. This dependency on the local grain structure may explain the seemingly random nature of pore formation in the HPDC process. This supposition is further supported by the grain size distributions in Fig. 6(c and d): a more uniform grain structure is produced under Optimized conditions, which is expected to result in a more homogeneous strain field, reducing scatter in pore size.

5. Conclusions

1. Scatter in tensile ductility is related to the size of large pores and non-metallic inclusions, for specimens produced under the observed conditions. A power-law relationship is established between the tensile ductility and the size of the largest defect on the fracture surface, which is irrespective of defect type.

2. We propose that these non-metallic inclusions form during the pyrolysis of commercial plunger lubricants in the shot chamber. This hypothesis is based on the chemical constituents of the inclusions, and is substantiated by shot chamber filling tests.

3. Scatter in tensile ductility is greatly reduced under Optimized conditions, due to a decrease in the maximum pore size from 1.32 mm to 0.37 mm. Our findings suggest that these large pores derive from dilatational strains introduced during semi-solid deformation. The apparent randomness of pore formation is thus attributed to the heterogeneous nature of the semi-solid network. Indeed, a more uniform grain structure is produced under Optimized conditions, which is expected to result in a more homogeneous strain field, reducing scatter in pore size.

Data availability

The raw/processed data required to reproduce these findings cannot be shared at this time due to legal or ethical reasons.

CRediT authorship contribution statement

Ewan Lordan: Conceptualization, Methodology, Validation, Formal
Appendix

To test the pyrolysis hypothesis, filling tests were performed by allowing the liquid to solidify in the shot chamber with the plunger at rest (Fig. A1). Fig. A1(b) shows a photograph of one of these filling test samples, with a close-up provided in Fig. A1(c). In Fig. A1(b), the location of the pouring hole is referred to by the phrase ‘inlet’; in the experiment, the plunger was in contact with the left-hand surface of the billet. As shown in Fig. A1(c), large films are observed on the surface of the billet. Furthermore, these films closely resemble the non-metallic inclusions shown in Fig. A1(b). Based on these results, it seems plausible that the aforementioned non-metallic inclusions form during the pyrolysis of commercial plunger lubricants. One might argue that the aforementioned elements are also found in commercial foundry fluxes; however, no flux was used during our experiments.

To ensure our findings are indeed representative of industry, fatigue specimens were machined from automotive components produced by HPDC. Various alloys were studied for samples produced by different suppliers. Crack initiating defects were identified via SEM fractography. Critical casting defects were identified as porosity, oxide films, and non-metallic inclusions—a representative BSE micrograph is shown in Fig. A1(a). Although, the Baseline process appears to be representative of industrial practice, oxides are more prominent in the automotive castings compared to Baseline samples. This is attributed to the increased geometric complexity of automotive components, and the use of sub-optimal melt handling procedures in commercial foundries (it is more difficult to control melt cleanliness in a commercial foundry than it is in a laboratory environment). As the suppliers cannot disclose their practices, we are unable to dismiss the use of foundry flux during casting.

![Fig. A1. Origin of non-metallic inclusions. (a) Representative micrograph of a non-metallic inclusion observed on the fracture surface of a fatigue specimen machined from an automotive component. (b) Photograph of a shot chamber filling test sample (inlet refers to the pouring hole). (c) Close up of the dashed region in (b).](image)

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