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1 **Combined deformation and solidification driven porosity formation** 2 **in Aluminium alloys**

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13 **Abstract**

14 In die-casting processes, the high cooling rates and pressures affect the alloy solidification
15 and deformation behaviour, and thereby impact the final mechanical properties of cast
16 components. In this study, isothermal semi-solid compression and subsequent cooling of
17 aluminium die-cast alloy specimens were characterised using fast synchrotron tomography.
18 This enabled the investigation and quantification of gas and shrinkage porosity evolution
19 during deformation and solidification. The analysis of the 4D images (3D plus time) revealed
20 two distinct mechanisms by which porosity formed; (i) deformation-induced growth due to
21 the enrichment of local hydrogen content by the advective hydrogen transport, as well as a
22 pressure drop in the dilatant shear bands, and (ii) diffusion-controlled growth during the
23 solidification. The rates of pore growth were quantified throughout the process, and a
24 Gaussian distribution function was found to represent the variation in the pore growth rate in
25 both regimes. Using a one-dimensional diffusion model for hydrogen pore growth, the
26 hydrogen flux required for driving pore growth during these regimes was estimated,
27 providing a new insight into the role of advective transport associated with the deformation in
28 the mushy region.

29 Keywords: die casting; gas and shrinkage porosity; dilatancy; synchrotron imaging; semi-
30 solid deformation

31 **1. Introduction**

32 The drive to improve fuel economy and reduce CO₂ emissions continues to incentivise the
33 development of low-cost lightweight high-strength cast alloys for automotive and other
34 transport applications ^[1]. These light alloys are required to possess excellent strength and

35 fatigue properties, together with good weldability and machinability, all at a low cost ^{[1][2]}.
36 These properties are heavily influenced by the presence of the microstructural features and
37 solidification defects like hot tear ^{[3][4]}, segregation ^{[5][6][7]}, and porosity ^{[8][9]}, which exist in
38 forms of (a) gas porosity ^[10], (b) shrinkage porosity ^[11], and shrinkage bands in twin-roll and
39 High pressure die castings (HPDC) ^[12]. The nucleation of these solidification defects can be
40 traced to the semi-solid state having relatively high solid fractions during the solidification. It
41 is known that at these higher solid fractions, a network of solid is formed and as a
42 consequence, the permeability of the mushy zone will decrease, resulting in difficulty in
43 further feeding of the liquid. Based on the amount of fractions of solid, the solid network has
44 been interpreted as a continuous solid skeleton ^{[13][14]} and cohesion-less granular solid
45 ^{[15][16][17]}. The thermo-mechanical response of this network under deformation is understood
46 to play a key role in the formation of defects.

47 Laboratory and synchrotron-based semi-solid deformation tests have been extensively carried
48 out by researchers to understand the thermo-mechanical behaviour of several aluminium
49 alloys, particularly binary Al-Cu alloys ^[18]. While tensile tests have been conducted to
50 determine the strength and ductility of the network ^[19], shear ^[20] and compression tests ^[18]
51 were used to study the rheology of the semi-solid. Tzimas et al. ^[18] reported semi-solid
52 compression tests of Al-4 wt.% Cu alloys that cover the effect of solid fraction, strain rate
53 and grain morphology and identified different factors affecting the flow resistance ^[18]. Kim et
54 al. ^[21], Kang et al. ^[22], and Kapranos et al. ^[23] conducted compression experiments to study
55 the rheological behaviour of various aluminium alloys at different solid fractions and strain
56 rates and reported liquid segregation and cracks at the edge of the specimens. The
57 development of advanced synchrotron experimentation via fast X-ray techniques have
58 allowed observations of the microstructural features during deformation *in situ* ^{[24][25]}. Kareh
59 et al. ^[26], and Cai et al. ^[27] reported *in situ* compression experiments and quantified the
60 granular motion and dilatancy at various imposed strains. The *in situ* studies confirmed the
61 role of volume dilation during tensile and compressive deformation of the semi-solid in
62 forming defects such as hot tears and shear bands ^[20]. Several theories of micro-mechanisms
63 of hot tear defect formation have been proposed and criteria for cracking have been
64 developed based on the experimental studies. ^{[28][29]}. However, most of these models do not
65 account for the local thermal history, formation of combined gas and shrinkage porosity, and
66 its effect on the initiation of hot cracks. Since the shrinkage and gas porosity are believed to
67 account for about 35% of the total defects in high pressure die cast components ^[30], a 3D,

68 real-time information of the defect formation is critical to develop predictive models for
69 porosity and cracks at both microscopic (size and shape) and macroscopic (location and
70 volume fraction) levels.

71 The diffusion driven growth of the gas micro-porosity based on the differential solubility of
72 hydrogen in the melt and the solid is well-known in the literature ^{[31] [32] [33]}. Early research on
73 gas porosity was focused on the quenching experiments and post-mortem observation of the
74 microstructures. Lee and Hunt ^[31] were the first to report the observation of porosity in Al-
75 Cu alloys in real-time using an X-ray temperature gradient stage, and the quantification of the
76 cooling rate on the pore radius and the volume fractions. Subsequently, experimental studies
77 on the hydrogen micro-porosity during directional solidification were reported by Arnberg
78 and Mathiessen ^[32], Liao et al. ^[33], and Lie et al. ^[34]. Catalina et al. ^[35] observed the change
79 in pore shape to ellipsoid when the pore is surrounded by the solutal layer ahead of the solid-
80 liquid interface and estimated the increase in growth rate during engulfment. Based on the X-
81 ray radiographic observations, several empirical models of pore growth, which account for
82 the influence of hydrogen diffusion, volumetric shrinkage ^[36] and presence of microstructural
83 features like intermetallics ^[37], have been reported. Likewise, the formation of shrinkage
84 porosity due to lack of feeding has been reported via several experimental studies. For e.g.,
85 Gourlay et al. ^{[16][20]} investigated the mechanism for the formation of shrinkage bands and Li
86 et al. ^[38] reported the influence of melt flow and externally solidified crystals on the
87 formation of defect bands in HPDC of AZ91D magnesium alloy.

88 However, the mechanisms of pore growth during deformation and purely convective
89 conditions, and the role of liquid flow in promoting porosity growth are neither reported nor
90 quantified, to the best of our knowledge. This information is critical to understand and
91 develop models to predict the size and location of porosity. In this study, compression of
92 semi-solid Al-Si-Cu die-cast alloys, with and without modified copper content, has been
93 performed to quantify the flow-driven pore nucleation and growth. The liquid fraction
94 distribution and the nucleated porosity in the dilatant bands were quantified at different strain
95 values during compression. A 1D diffusion-controlled gas porosity model was used to
96 quantify the flux required for pore growth, which provided insights into the propensity of
97 advective hydrogen transport in enhancing the growth of hydrogen pores. In what follows, we
98 present the experimental methodology, analysis and quantification of the 4D (3D+time) data
99 characterising the nature of deformation-induced and solidification-driven pore growth.

100

101 2. Materials and Methods

102 Die-cast aluminium alloy ADC12 and a Modified ADC12 alloy (will be termed as MADC12
103 henceforth) with an additional 10 wt.% copper were used in the study. The modified alloy
104 was used to achieve an enhanced contrast between the primary phase and the inter-dendritic
105 liquid. A secondary objective of modifying the alloy was to investigate the influence of the
106 change in freezing range on hydrogen pore growth, and thus the propensity of cracking. The
107 chemical composition of the major alloying elements (Table 1) was assessed using
108 inductively-coupled plasma atomic emission spectroscopy (ICP-AES) technique.

109 **Table 1: Chemical assay of the alloys under study**

Sample	Cu (wt.%)	Fe (wt.%)	Si (wt.%)
ADC12	1.82	0.74	11.9
Modified ADC12	10.43	0.80	10.5

110

111 2.1 Sample preparation

112 Calculated amounts of ADC12 alloy and 99.99% copper were melted in a coreless induction
113 furnace fitted with a pure graphite crucible lined with magnesite refractory. The molten alloy
114 was continuously stirred for 2 minutes with a graphite rod to ensure that the alloy
115 composition is uniform. Then the metal was poured into a cast iron die of dimension $250 \times$
116 $50 \times 40 \text{ mm}^3$ and allowed to cool in air. The ADC12 ingots and MADC12 alloy ingots were
117 then machined into 3 mm diameter cylinders using wire electro discharge machining. The
118 length of the final samples was $5.4 \pm 0.3 \text{ mm}$. The Scheil solidification model in Thermo-
119 Calc®^[39] was used to obtain the freezing range for determining the experimental conditions,
120 and is shown in figure 1. The **estimation of solidification range by Thermo-Calc ©** were
121 validated by conducting bulk cooling experiments as shown in inset of figure 1. Since
122 ADC12 has a larger freezing range, the hot-cracking propensity is expected to be higher.

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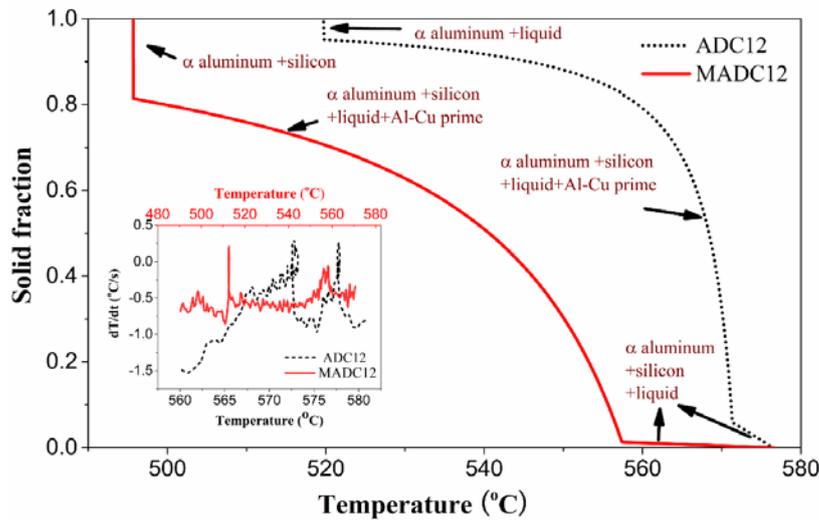
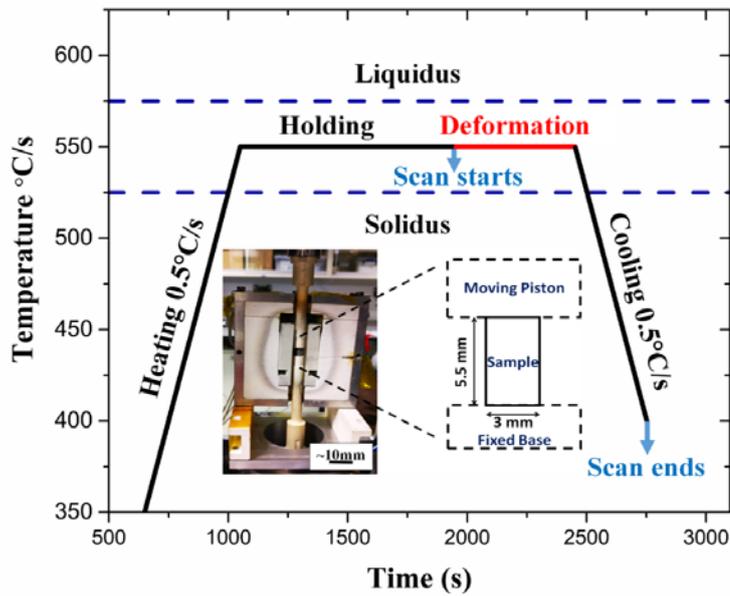


Fig. 1—Solid fraction vs. temperature for MADC12 and ADC12 alloys, calculated using the Thermo-Calc® Scheil solidification module. The inset shows the cooling rates of the bulk samples being cooled at 0.5 °C/s, with the lower X-axis indicating the temperatures for ADC12 alloy and the upper X-axis for MADC12 alloy.

2.2 Semi-solid compression experiments

The semi-solid compression experiments were conducted at I12 beamline of the Diamond Light Source^[40] (beamtime reference - EE16188). A PID-controlled resistance furnace fitted with an X-ray transparent window is mounted on a custom-made mechanical rig for the semi-solid compression experiments. Details of the setup can be found in^{[27][41]}. The mechanical rig used was specifically designed to conduct *in situ* deformation and has high precision loading control (as low as 100 nm/s) and force measurement (least count 0.1 N). A schematic of the experimental set-up is shown in figure 2. The specimen was placed at the centre of alumina shackles as shown in the figure 2. A preload of approximately 7 N was applied to hold the sample in place during compression experiments. The loading ram was adjusted to maintain the preload during the experiment to account for thermal expansion during heating. The thermal cycle followed during the experimentation along with the sample arrangement are presented in figure 2. A loading rate of 5 μm/s and the solid fraction of 75%±5% were maintained during the compression experiments. The fraction of solid was determined by the temperature from Scheil modelling as well as from the *in situ* images.



143
 144 Fig. 2—Thermal cycle followed during the semi-solid compression test. The inset shows the
 145 sample mounted inside the furnace.

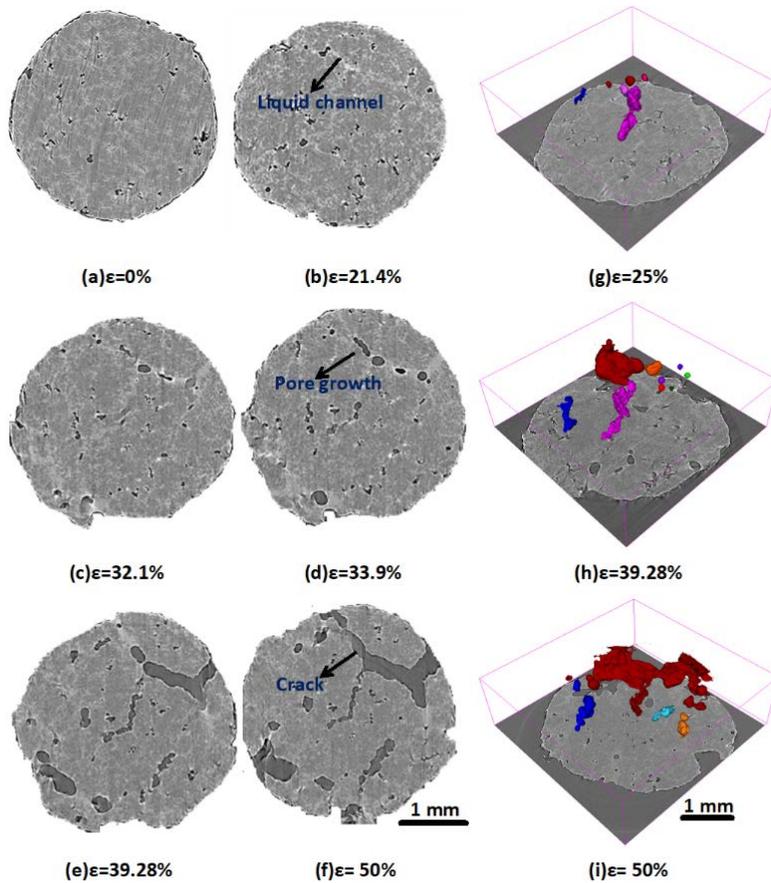
146 **2.3 Data acquisition and image processing**

147 A monochromatic X-ray beam with an energy of 53 keV was used in the experiment. During
 148 the deformation, a set of 38 tomograms were acquired using a PCO.Edge camera coupled
 149 with I12's camera module 3, which corresponds to a field of view of 8 mm × 6 mm^[40]. A
 150 total of 600 projections were taken for every 180° rotation of the sample for each tomogram,
 151 obtaining a voxel size of 3.2 μm. The exposure time was 32 ms per projection, with the total
 152 time of each scan being 19.2 s. Each scan was taken continuously without any additional
 153 sample rotation, which allowed for the continuous data collection. The 3D scan was
 154 reconstructed using filtered back-projection to produce an 1885 × 1885 × 2149 voxel volume.
 155 The image was filtered using anisotropic and 3D median filters to remove noises. The filtered
 156 image was segmented using trainable weka segmentation plugin^[42] in Fiji ImageJ^[43], which
 157 makes use of machine learning tools.

158 **3. Results and Discussions**

159 The transverse slices of the MADC12 sample at an approximate height of 600 μm from the
 160 bottom slice and at 6 different strain values are shown in figure 3 (a-f). In the figure, the dark
 161 grey represents the porosity and the lighter grey indicates the inter-dendritic liquid, while the
 162 intermediate grey colour corresponds to the primary solid phase. The longitudinal strain

163 values indicated were measured by dividing the instantaneous length of the sample by the
 164 initial length. The sequence of the mechanisms leading to the formation of a crack is
 165 described as follows.

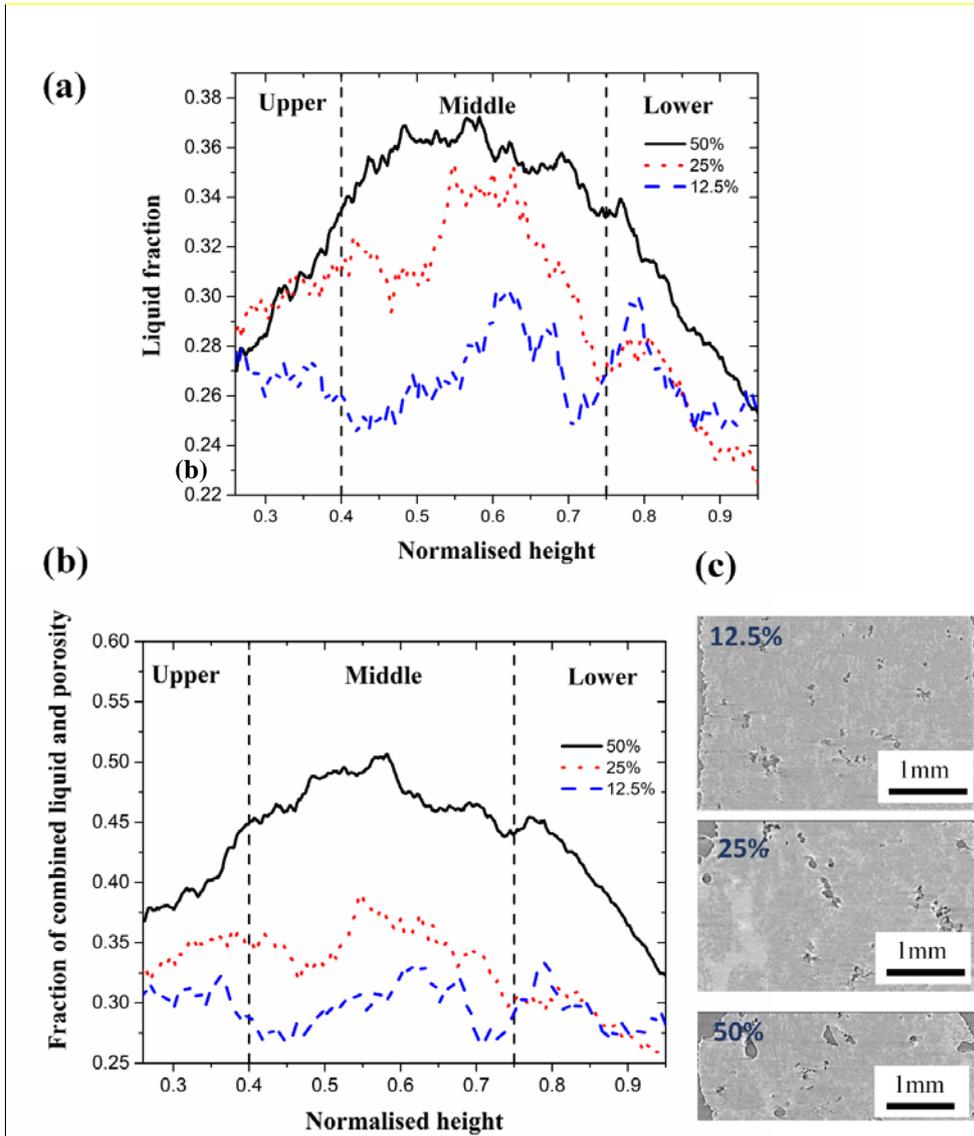


166
 167 Fig. 3— (a-f) Transverse slices of the MADC12 sample (75% fraction solid) at a height of
 168 ~600 μm from the sample bottom at different strains, showing the formation and growth of
 169 gas porosity and the subsequent development of a hot tear (g-i) 3D rendered images at
 170 equivalent time instances, with each colour representing a contiguous segment of porosity.

171 3.1.1 Formation of liquid channels

172 The inter-dendritic liquid, which was distributed uniformly ($\sim 25 \pm 5\%$) (figure 3(a)) before the
 173 compression commenced, formed several isolated liquid channels on the application of
 174 compression load (one such channel at $\epsilon=21.4\%$ is marked in figure 3(b)). The measured
 175 values of liquid fraction at different strains along the loading axis (figure 4(a)), which shows
 176 the localization of liquid channels in the middle of sample. Note that the height of sample for

177 each area liquid fraction was normalized with respect the total height of the sample at that
 178 instant of deformation. The increase in the fraction is due to the increased area at the middle
 179 of the sample coupled with the development of liquid channels due to dilatancy. These liquid
 180 pockets were later observed to be the location of nucleation of gas micro-porosity during
 181 deformation as well as the cooling stages.



182
 183 Fig. 4—Development of the (a) liquid (b) combined liquid and pore fraction along the
 184 loading axis at different strain values. (c) Longitudinal slices at different strain values
 185 indicating the growth of liquid channel

186 **3.1.2 Pore nucleation, growth and coalescence**

187 The nucleation of new gas pores was observed in the liquid channels at various strain values
188 throughout the deformation stage (figure 3(b)). The pores grew consistently while the mush
189 deformed with expanding liquid channels. Plausibly, the feeding of the liquid resulted in a
190 convective influx of hydrogen into the liquid pockets, thereby increasing the overall local
191 hydrogen concentration. This is hypothesized based on the observation that the pores were
192 predominantly spherical during the deformation regime, indicating a diffusion-controlled gas
193 pore growth. Furthermore, a secondary influence on pore growth was the pressure drop in the
194 dilatant bands, which can also lead to pore volume dilation. The pores retained spherical
195 shape until they encountered a solid (figure 3d). The phenomenon was also observed and
196 reported by Cai et al., [27] who had defined different stages of dilatancy controlled shrinkage
197 growth, which eventually led to cracking. The values of combined area of liquid fraction and
198 porosity at different strains along the loading axis were determined and shown in figure 4(b).
199 By comparing figure 4(a) and 4(b) it is clear, that the pore growth due to volume dilation
200 picks up after 25% strain and most of these voids were concentrated around the middle of the
201 sample. Longitudinal slices shown in figure 4(c), shows the ‘middle’ region indicated in the
202 figure 4(a) and (b), in which the localization of liquid channel and voids can be seen. **It**
203 **should be noted that, in earlier studies by Stefanescu [44] and Khalajzadeh [45], a shrinkage**
204 **induced flow term, introduced and its role on shrinkage growth had been discussed. This flow**
205 **term is analogous to the dilatancy induced flow discussed in this study.**

206
207 **3.2 Regimes of gas porosity growth: (a) deformation and (b) solidification**

208 The gas pores which did not coalesce and grew independently during both deformation and
209 solidification stages were tracked from the tomograms obtained from the *in situ* experiments.
210 The radius of these isolated pores at various strain values were measured and data of 2 typical
211 pores are represented in figures 5 for the MADC12 alloy. It is evident that the growth
212 behaviour of the pores altered noticeably at the end of the deformation (when the cooling
213 initiated). Based on the experimental observations, two regimes of pore growth were defined,
214 namely, (a) deformation induced and (b) solidification controlled.

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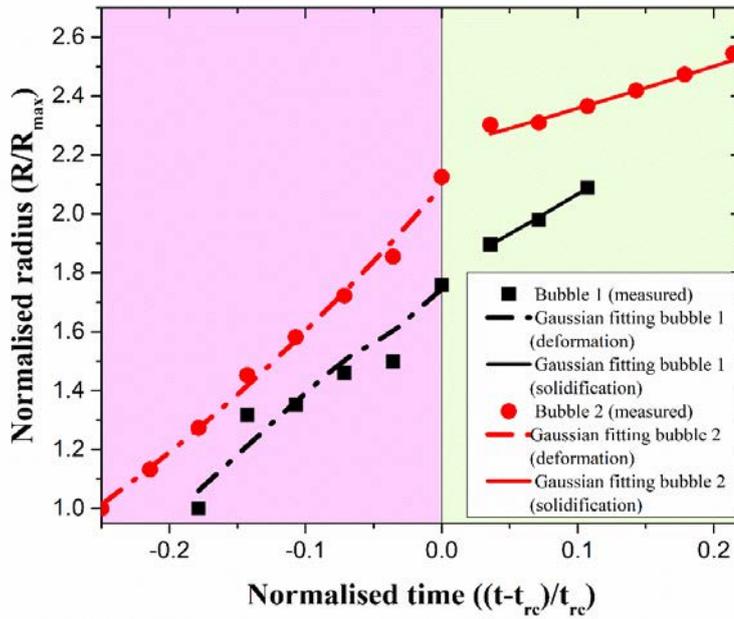
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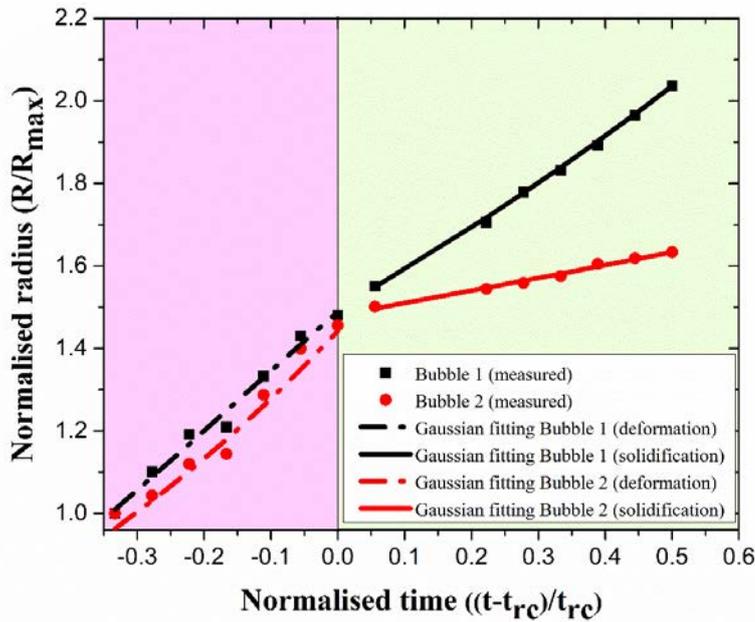
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216
 217 Fig. 5—Evolution of the gas micro-porosity with time during deformation and cooling
 218 (solidification) stages for MADC12 alloy. Here R_{max} and t_{rc} represents the maximum radius
 219 (end of solidification) and time instant at which the regime changes respectively



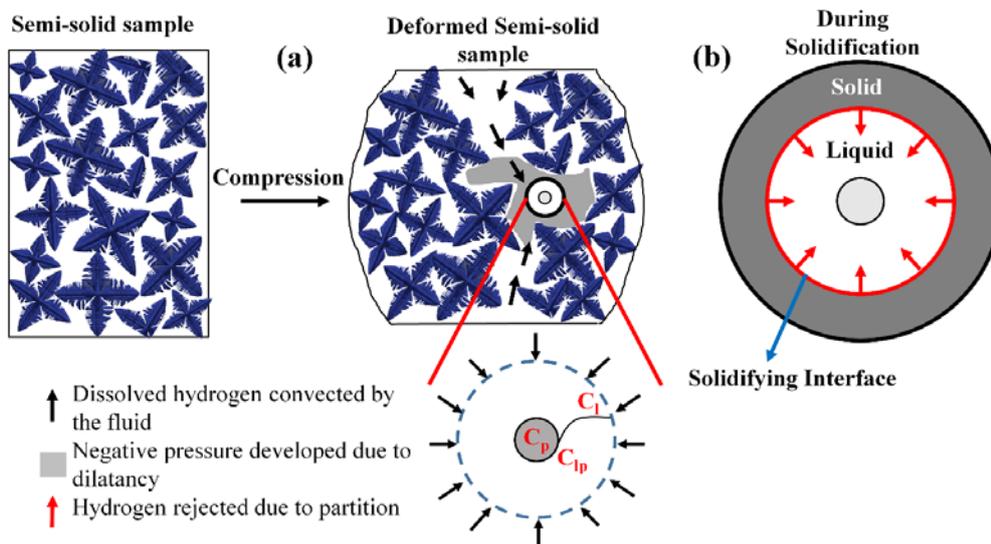
220
 221 Fig. 6—Evolution of gas micro-porosity with time during deformation and cooling
 222 (solidification) stages for ADC12 alloy.

223
 224 Recently, Sun et al., modelled the diffusion-controlled gas pore growth in aluminium melt
 225 using a Gaussian distribution function [37]. Following their work, the pore growth in two
 226 regimes of the present data was fit using a Gaussian distribution function as given in equation
 227 1.

$$228 \quad r = \frac{A}{t_{SD} \times \sqrt{2\pi}} \times \exp\left(-2 \frac{(t - t_{mean})^2}{t_{SD}^2}\right) \quad [1]$$

229 where, A is a constant corresponding to the amplitude of the Gaussian distribution, and t_{mean}
 230 and t_{SD} are mean and standard deviation of the Gaussian function. The fitted curves are
 231 shown with the experimental observations in figures 5 and 6 and the adjusted R^2 value
 232 indicate a reasonably good fit.

233 The proposed hypothesis of a combined deformation and solidification-driven hydrogen
 234 micro-porosity is represented schematically in figure 7. The regions within the circles
 235 essentially show a representative elementary volume (REV) surrounding a gas pore. Figure
 236 7(a) shows the deformation-driven flow of the liquid, which fed the hydrogen into the pore by
 237 convective transport, while 7(b) shows the rejection of hydrogen during the solidification of
 238 neighbouring solid structures.



239
 240 Fig. 7— Schematic of the proposed pore growth mechanism during (a) deformation and (b)
 241 solidification. The black arrows in (a) represents the liquid flow associated with dilatancy.

242 The black arrows in (a) and red arrows in (b) represent hydrogen flux into the representative
243 volume around a gas pore.

244 3.3. Numerical evaluation of advective hydrogen influx

245 Since the isolated pores show a distinct growth behaviour during deformation as compared to
246 solidification, an understanding of amount of hydrogen influx carried by the liquid would be
247 useful in correlating the flow-induced growth under different processing environments such
248 as twin roll casting, HPDC etc. A calculation of 1D diffusion-driven hydrogen gas pore
249 growth was performed using a numerical model discussed in [46]. It was assumed that a
250 constant value of hydrogen concentration existed at the boundary of the domain (figure 7),
251 which represents the amount of hydrogen brought in by the inter-dendritic flow. Using the
252 pore growth model, the necessary hydrogen concentration at the boundary that resulted in an
253 equivalent growth obtained from the experiments was determined. Note that this analysis
254 only accounts for the growth of the spherical pore under the influence of increased hydrogen
255 influx due to liquid flow.

256 A pre-existing hydrogen gas pore of radius R in liquid supersaturated with hydrogen was
257 considered, which grows instantaneously due to diffusion of hydrogen into the pore. The size
258 of the REV (with radius R_{domain}) was assumed to be larger compared to the diffusion
259 boundary layer. By balancing the rate of mass increase inside the pore, the influx of
260 concentration at boundary, and the flow due to the movement of the boundary itself, the
261 growth rate of the liquid-pore interface is evaluated as follows:

$$262 \quad V_{int} = \frac{1}{(C_p - C_{lp} - C_d)} D_l \frac{\partial C_l}{\partial r} \quad [2]$$

263 where C_b , C_p , C_{lp} , C_d are the hydrogen concentration (in cc/100 g of Al) in the far field,
264 inside the pore, at the interface and an additional term arising due to the liquid-pore interface
265 movement respectively, and are functions of the pore radius and reference pressure (details in
266 supplementary information 2). The details of the individual terms are given in supplementary
267 information 2. D_l is the mass diffusivity of hydrogen in aluminium. In this study, for the
268 calculation of C_p and C_{lp} the reference pressure for the liquid was set at 1 atm. The surface
269 tension and the mass diffusivity values were taken from Sasikumar et al [47] and Lee et al [31].
270 [46]. The equivalent radius, measured from the tomographic scans, was assumed to be the
271 initial radius of the sample. A fixed concentration input at the far-field liquid that drives the
272 influx of hydrogen can be obtained as follows:

$$C_l = C_{lp} + \frac{V_{int}(C_p - C_{lp} - C_d)}{D_l} R$$

274 [3]

275 Using the above relation, a concentration change created by the deformation-induced flow
 276 was estimated and is reported for a few pore radii in Table 2. The model was used to
 277 calculate the input concentration for a range of input radii (12.47-36.01 μm for MADC12 and
 278 69.95-131.74 μm for ADC12) and the corresponding rate of change of the concentration at
 279 the REV interface was calculated and found to be of the order $\sim 10^{-5}$.

280 Table 2: The concentration input at domain boundary estimated from the model

Alloy	Radius of the gas pore (μm) in the two regimes			Avg. concentration input (cc/100g Al) between two scans	
	Pore	Deformation	Solidification	Deformation	Solidification
MADC12	1	29.8	68.8	2.6×10^{-4}	1.9×10^{-5}
	2	37.9	70.4	6.3×10^{-4}	1.5×10^{-5}
ADC12	1	69.9	119.3	8.0×10^{-4}	1.1×10^{-4}
	2	77.0	124.5	4.8×10^{-4}	5.2×10^{-5}

281

282 Using mass conservation, this rate of change of concentration at the interface is related to the
 283 convective flux given by the equation 4. Note that we have assumed that the velocity of the
 284 liquid flow is of the same order as the deformation rate (V)

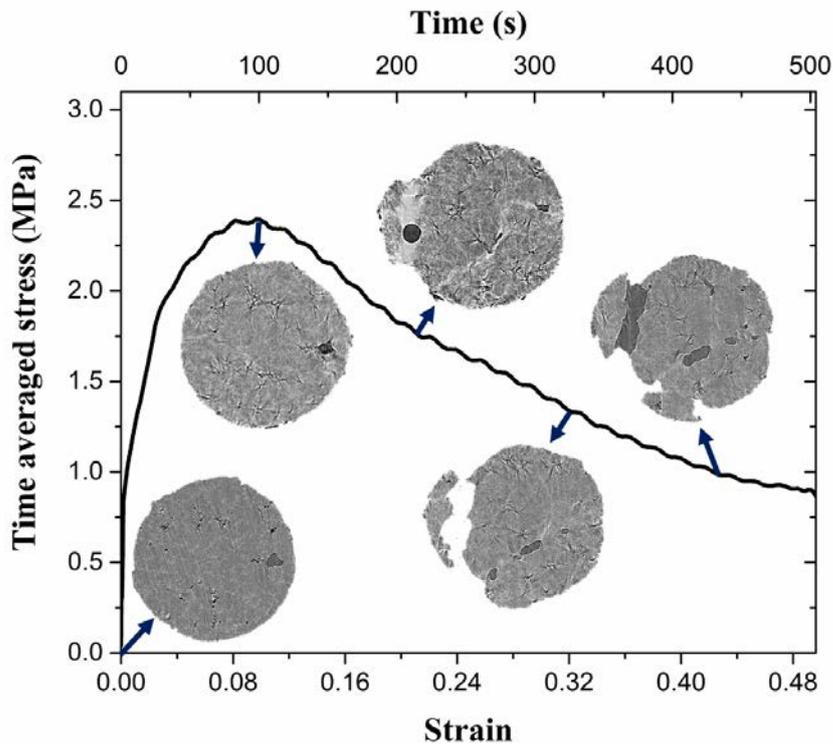
$$285 \frac{\partial C}{\partial t} = V \frac{\partial C}{\partial x} \quad [4]$$

286 It can be seen that by increasing the velocity of deformation, the concentration build-up at
 287 REV increases, for a given far-field concentration. This increase in the concentration gradient
 288 between the REV boundary and the pore-liquid interface drives the pore growth. This
 289 information is helpful in evaluating flow-driven hydrogen concentration build-up and
 290 eventual porosity, particularly for high-pressure die casting environments, where flow rates
 291 are very high ($\sim 20\text{-}60$ m/s) [2].

292 3. Stress-strain behaviour during compression

293 The occurrence of dilatancy is ascertained with the help of flow stress vs. strain data obtained
 294 from the experiment. The normalised stress vs. strain curve with the insets in the figure
 295 indicating the transverse slice at every 10% strain from the semi-solid compression is shown
 296 in the figure 8. Note that the stress is calculated as the ratio of the applied load to the initial
 297 sample area. The graph shows characteristic peak at a strain value of ~ 0.1 which indicated

298 that the stress developed due to the pinning of the grains was being relieved by the
299 development of liquid channels due to dilatancy (inset at 20% strain). Further dilation of the
300 liquid channel and nucleation and growth of porosity resulted in a continuous decrease of the
301 load, and eventual formation of a hot-tear at around 0.32 strain.



302
303 Fig. 8—Stress-Strain curve for MADC12 alloy. Insets display the transverse slice of the
304 specimen at the indicated strain values in the range of 0-50%.

305 **Conclusions:**

306 Ultrafast synchrotron X-ray imaging was utilized to observe the porosity growth in 2
307 different Al-Si-Cu die-cast alloy specimens at $\sim 75 \pm 5\%$ solid fraction during compression and
308 solidification conditions. In the present work, we have proposed a mechanism of flow-driven
309 hydrogen pore growth which is critical for predicting final porosity in high-pressure die-cast
310 components. The following conclusions were drawn from the study.

- 311 1. Two distinct regimes of hydrogen pore growth, namely deformation-induced flow driven
312 growth and solidification controlled growth were observed and quantified. The growth
313 rates were determined by measuring the equivalent pore radius from *in situ* experiments.

314 The measured growth rate values showed a clear and abrupt change at the end of
315 deformation regime, and were also found to agree with the trends reported in the literature
316 using Gaussian distribution functions.

317 2. The role of dilatancy in deformation induced pore growth via advective hydrogen
318 transport was ascertained using quantification of localized liquid channels and the stress-
319 strain behaviour of the semi-solid under compressive loading.

320 3. The hydrogen concentration influx during deformation induced flow that closely
321 represents the observed experimental trend was estimated using a 1D diffusion model.
322 This revealed a hydrogen concentration boundary condition, and can be further used to
323 correlate flow-driven hydrogen influx for different processing routes.

324

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422 **Figure Captions**

423 Fig. 1—Solid fraction vs. temperature for MADC12 and ADC12 alloys, calculated using the
424 Thermo-Calc® Scheil solidification module. The inset shows the cooling curves of the bulk
425 samples being cooled at 0.5 °C/s, with the lower X-axis indicating the temperatures for
426 ADC12 alloy and the upper X-axis for MADC12 alloy.

427

428 Fig. 2—Thermal cycle followed during the semi-solid compression test. The inset shows the
429 sample mounted inside the furnace.

430

431 Fig. 3— (a-f) Transverse slices of the MADC12 sample (75% fraction solid) at a height of
432 ~600 µm from the sample bottom at different strains, showing the formation and growth of
433 gas porosity and the subsequent development of a hot tear (g-i) 3D rendered images at
434 equivalent time instances, with each colour representing a contiguous segment of porosity.

435

436 Fig. 4—Development of the (a) liquid (b) combined liquid and pore fraction along the
437 loading axis at different strain values. (c) Longitudinal slices at different strain values
438 indicating the growth of liquid channel

439

440 Fig. 5—Evolution of the gas micro-porosity with time during deformation and cooling
441 (solidification) stages for MADC12 alloy. Here R_{max} and t_{rc} represents the maximum radius
442 (end of solidification) and time instant at which the regime changes respectively

443

444 Fig. 6—Evolution of gas micro-porosity with time during deformation and cooling
445 (solidification) stages for ADC12 alloy.

446

447 Fig. 7— Schematic of the proposed pore growth mechanism during (a) deformation and (b)
448 solidification. The black arrows in (a) represents the liquid flow associated with dilatancy.
449 The black arrows in (a) and red arrows in (b) represent hydrogen flux into the representative
450 volume around a gas pore.

451

452 Fig. 8—Stress-Strain curve for MADC12 alloy. Insets display the transverse slice of the
453 specimen at the indicated strain values in the range of 0-50%.