Carbon Fibre Reinforced Aluminium and Magnesium Materials for Automotive Structures

Towards Affordable, Closed-Loop Recyclable Future Low Carbon Vehicle Structures

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Abstract

This report outlines the work completed in the 2\textsuperscript{nd} year of this PhD research project, contributing to the TARF-LCV project (Towards Affordable, Closed-Loop Recyclable Future Low Carbon Vehicle Structures).

The majority of this report takes the form of a journal paper, the aim of which was to determine whether the addition of silicon to a carbon fibre reinforced aluminium composite material would inhibit the formation of unwanted aluminium carbide at the matrix/reinforcement interface.

Other work completed can be found in the appendix, along with a thesis plan and a proposed plan for future work.
The Effect of Silicon Content on the Reaction of Carbon Fibre in Molten Aluminium Alloys


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Keywords: carbon fibre, aluminium, metal matrix composites

1. Introduction

Carbon fibre (CF) reinforced aluminium (Al) has desirable properties for the automotive industry, such as high specific stiffness and strength [1, 2]. Volume fraction of CFs as low as 5% can markedly improve wear resistant of the composite material compared with unreinforced Al [3].

Discontinuous fibres are of particular interest due to their versatility, relatively short cycle times, reduced anisotropy and low wastage compared with woven fabrics [4-7]. Recent activities in CF recycling [8, 9] has demonstrated cost effective routes for recovering high quality fibres in discontinuous form. This provides further motivation to develop processes to exploit low cost fibres for high performance applications.
However, the composite material does not perform as well as predicted by rule of mixtures (ROM) during mechanical testing, due to problems at the matrix/reinforcement interface [1]. This is due to poor wetting between the matrix and reinforcement, and the formation of brittle aluminium carbide (Al₄C₃), both of which degrade the mechanical properties of the resulting composite material [1, 2, 10-13]. CF coatings are often employed to either increase the wettability between the molten alloy and fibre, or to act as a barrier to reduce prevent interfacial reactions [14]. However, the use of coatings necessitates an additional processing step, adding cost to the finished material.

The Al carbide formed at the interface, Al₄C₃, is hydrophilic, so in moist environments its presence will increase the rate of crack growth [15]. Length of time at processing temperature can also increase the amount of interfacial carbide present; Li et al [1] showed that after 60 hours of heat treatment at 600°C, CF/Al had 10% reacted carbon (by volume) and a UTS of 67.2MP, compared with 1% reacted carbon and 172MPa UTS for the as-processed composite material. Interfacial properties are critical to the overall properties of metal matrix composites (MMCs) [15-19]. It is therefore crucial to reduce these interfacial reactions, while improving the wettability between the matrix and reinforcement.

Alloying of the matrix material has been suggested as an alternative way to reduce interfacial reactions [2, 18, 20]. Alloying Al with silicon (Si) results in higher melt fluidity, low solidification shrinkage and improved resistance to corrosion, but reduces machineability due to the hardness of the Si phase [21]. Revzin et al [10] and Pelleg et al [11] alloyed Al matrices with 0.5 atomic % Si and saw a reduction in interfacial carbide formation. Pelleg additionally noted that the interfacial carbide in the system without silicon reduced the fibre diameter by up to 3μm, whereas no fibre degradation
was noted for the system with silicon. However, in both of these cases, the interfacial aluminium carbide was only slightly reduced. Chernyshova et al [13] reduced Al$_4$C$_3$ content in CF/Al composites by over 40% by alloying the matrix with 12wt%Si.

The aim of this paper is to investigate the interfacial reactions present in CF/Al, alloyed with varying amounts of Si, fabricated by gas pressure infiltration. This work intends to determine how much Si in the Al matrix is required to suppress the formation of unwanted Al$_4$C$_3$, and instead facilitate the formation of SiC.

Thermodynamic modelling has been considered to calculate the minimum amount of Si required, and this will be validated experimentally.
2. Thermodynamic Theory

The modelling approach considers the reaction of carbon (graphite) in a binary Al-Si alloy. In the present calculation only 3 solid phases are considered: C (graphite), SiC (silicon carbide) and Al₄C₃ (aluminium carbide); the formation of ternary carbides is ignored. Any deviation from the stoichiometry and any dissolution of Al or Si in these solid phases are neglected, as is the solubility of carbon in the liquid alloy.

Two chemical processes (Equation 1 and Equation 2) are initially considered to take place:

Equation 1

\[ Si_{\text{alloy}} + C = SiC \]

Equation 2

\[ Al_{\text{alloy}} + \frac{3}{4} \cdot C = \frac{1}{4} Al_4C_3 \]

Knowing the standard Gibbs energy for the 3 solid phases as a function of temperature and for the liquid alloy (as function of T and mole fraction of Si in the liquid, \( x_{Si} \)) the Gibbs energy changes for can be determined the above reactions [22, 23]. Reaction of the C-fibre with the liquid alloy (to form either SiC or Al₄C₃ in Equation 1 and Equation 2) will occur if \( \Delta G \leq 0 \), defining the boundaries between the (L+C) and (L + reaction product) two-phase regions.
The equilibrium line (blue line) separating two, two-phase regions: liquid Al–Si alloy in contact with solid $\text{Al}_4\text{C}_3$ and with solid SiC coating the original C fibres.

Calculations show (presented for the quasi-binary system in Figure 1) that at high temperatures ($>1700$°C) the two critical curves defining Equation 1 and Equation 2 do not overlap. In this case, the spaces of the ternary diagram not occupied by the (L + SiC) and (L + $\text{Al}_4\text{C}_3$) regions are stable (L + C) regions. However, for temperatures relevant to this study (in the range 700 – 800°C), these critical boundaries overlap, giving rise to the following reaction:

\[
\frac{3}{4} \cdot \text{SiC} + \text{Al}_{\text{alloy}} = \frac{1}{4} \text{Al}_4\text{C}_3 + \frac{3}{4} \cdot \text{Si}_{\text{alloy}}
\]
This reaction defines the condition where a C-fibre initially reacted to SiC will be converted into Al₄C₃, if \( \Delta_{R_0} G \leq 0 \). If this condition is fulfilled, the two-phase region (L + Al₄C₃) will be stable (supposing \( \Delta_{R_2} G \leq 0 \) is also fulfilled). If this condition is not fulfilled, the two-phase region (L + SiC) will be stable (supposing \( \Delta_{R_1} G \leq 0 \) is also fulfilled).

The results of calculations for the critical mole fractions of Si for reaction 3, in the temperature interval 933 - 1173 K, using Equation 4, are shown in Figure 1, where the blue line defining the boundary is shown overlapping the Al-rich end of the Al-Si phase diagram.

**Equation 4**

\[
x_{Si,cr}^{Al_4C_3/SiC} = \exp\left(-1.553 - \frac{1113}{T}\right)
\]

If reaction of the carbon fibres in the MMCs investigated in this study cannot be avoided, then the preferred reaction product is SiC, rather than the very brittle and highly hygroscopic Al₄C₃. In this study, composite fabrication and heat treatment were performed at 750 °C (1022 K) and hence \( x_{Si,cr}^{Al_4C_3/SiC} = 0.071 \), corresponding to 7.5 mass % of Si in the liquid solution. If the Si-content of the liquid alloy is higher than 7.5 mass %, then the C-fibres at 750°C are expected to react to SiC rather than to Al₄C₃.
3. Experimental

Commercially pure Al (99.7wt%Al) and Al–Si alloys with 3, 6, 9 and 12wt%Si were used as matrix materials. Al–Si alloys were melted and mixed from 99.7wt%Al and 99.9wt%Si.

Carbon fibre was in the form of a non-woven random mat manufactured from 24K Sigrafil C, supplied by SGL. Carbon tows were cut and filamentised using a laboratory scale chopping head [24] attached to a 6-axis robot. Fibres were sprayed to create a 400gsm random mat, which was needle punched together to avoid using an adhesive binder. The CF mat was stamped in Φ20mm circles and heated to 500°C for 1 hour to remove any sizing or impurities.

<table>
<thead>
<tr>
<th>Matrix alloy</th>
<th>C.P.Al</th>
<th>Al–3Si</th>
<th>Al–6Si</th>
<th>Al–9Si</th>
<th>Al–12Si</th>
</tr>
</thead>
<tbody>
<tr>
<td>Heat treatment at 750°C (hrs)</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>X</td>
<td>X</td>
<td>X</td>
<td>X</td>
<td>X</td>
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<tr>
<td>2</td>
<td></td>
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<td>X</td>
<td>X</td>
<td>X</td>
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<td>24</td>
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<td>X</td>
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<tr>
<td>48</td>
<td>X</td>
<td></td>
<td></td>
<td></td>
<td>X</td>
</tr>
</tbody>
</table>

Table 1

*Matrix of experiments, X indicates a fabricated composite material*
CF/Al composite materials were fabricated by gas pressure infiltration (GPI). 35g of Al and 0.77g of CF were placed in an alumina crucible, heated to 750°C under vacuum and held at temperature for 90 minutes. 25 bar of Ar gas pressure was applied for 30 minutes at temperature, and subsequently until the system had cooled (see Figure 2 for schematic of GPI set-up).

Samples were then mounted using Conductomount (phenolic moulding compound with graphite fibres) and polished without water (to avoid reaction between aluminium carbide and water [15]) and studied under the scanning electron microscope (SEM) and optical microscope (OM). X-ray diffraction (XRD) was used to identify the elements present at the interface. Volume fraction (VF) of the composite materials was determined using ImageJ software.
Figure 2
Pressure vessel set-up, showing (a) termocouple shaft, (b) vacuum port, (c) gas inlet, (d) gas outlet, (e) stainless steel pressure vessel, (f) heating elements, (g) alumina crucible, (h) aluminium ingot, (i) carbon fibre
4. Results

Figure 3 shows all five composite materials, with varying amounts of Si in the Al matrix. The fibres are evenly distributed, some agglomeration of fibres can be seen in Figure 3. Porosity only occurs between the occasional bundles of fibres. The Al–Si matrix alloy is as expected, with needle shaped Si within the Al. The fibre/matrix interfaces are clean, compared with reacted fibres shown in Figure 5, with no signs of reaction products after 30 minutes processing time at 750°C. The VF of the composite materials is in the region of 9 – 15% (see Table 2).

Figure 4 shows how subsequent heat treatments have caused the fibres and matrix to react over time. After 1 hour at 750°C, the composite material looks similar to the as-processed material, with clean interfaces and fully formed fibres (Figure 4a). After 4 hours of heat treatment, reaction products are beginning to form at the fibre/matrix interface. These appear as dark rings around the outside of the CFs (Figure 4b). After 8 hours of heat treatment, the CFs appear hollow with increased presence of interfacial reactions (Figure 4c and d). Further to this, Figure 6 shows an x-ray mapping image of before and after 8 hours of heat treatment. Carbon can clearly be identified in the as-processed composite (shown in red), but has almost disappeared from the heat treated composite. After 24 hours, CFs in their original form are no longer visible.

<table>
<thead>
<tr>
<th>Composite matrix</th>
<th>Actual Si content (ma%)</th>
<th>VF of CFs (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C.P. Al</td>
<td>-</td>
<td>14.6</td>
</tr>
<tr>
<td>Al–3Si</td>
<td>2.89</td>
<td>9.5</td>
</tr>
<tr>
<td>Al–6Si</td>
<td>5.73</td>
<td>11.4</td>
</tr>
<tr>
<td>Al–9Si</td>
<td>8.95</td>
<td>10.6</td>
</tr>
<tr>
<td>Al–12Si</td>
<td>12.18</td>
<td>12.3</td>
</tr>
</tbody>
</table>

Table 2

*Si content in Al matrix alloys, and VFs of CF in resulting composite materials*
Figure 3

SEM images using BSE detector at 500x magnification of (a) CF/Al–12Si, (b) CF/Al–9Si, (c) CF/Al–6Si, (d) CF/Al–3Si, (e) CF/C.P.Al, all as processed
Figure 4
OM images of CF/Al–9Si (a) 1 hour 750°C, (b) 4 hours 750°C, (c) 8 hours 750°C, (d) 24 hours 750°C

Figure 5
OM images of (a) CF/Al–12Si and (b) CF/C.P.AI after 24 hours 750°C
Figure 6
X-ray mapping using INCA software of CF/Al–9Si at 2000x magnification (a) as processed, (b) 8 hours 750°C, where green, blue and red represent Al, Si and C, respectively.

Si in the Al matrix above 7.5wt% should theoretically facilitate the formation of SiC at the matrix/reinforcement interface, rather than Al₄C₃. Figure 7 shows the XRD data obtained from CF reinforced C.P.Al and Al–12Si after 48 hours at 750°C. XRD has been used to identify the elements and compounds present in the composite material. Al₄C₃ is present in both samples; in fact most Al₄C₃ diffraction peaks are larger from the sample containing Si. Figure 10 compares CF reinforced Al–12Si and C.P.Al, heat treated for 48 hours; each fibre is degraded and surrounded by a layer of Al₄C₃. Little difference can be seen between the 2 materials, supporting the XRD data and x-ray mapping.
Figure 7
XRD data for (a) CF/C.P.Al and (b) CF/Al–12Si after 48 hours 750°C, black arrow indicates SiC peak
5. Discussion

All MMCs should exhibit a strong matrix/reinforcement interfacial bond, to ensure that load is effectively transferred from the matrix to the reinforcement. The carbide Al₄C₃ forms at the interface between Al and CFs at processing temperature, which can initiate the formation of cracks and degrades the structural integrity of the fibres. Thermodynamic analysis (see section 2) stated that the addition of Si to the matrix above 7.5wt% would preferentially facilitate the formation of SiC at the interface, rather than Al₄C₃, however the presence of Al₄C₃ detected by the XRD in samples with and without Si in the matrix suggests that this may not be the case.

The addition of Si to the matrix does have some benefits. Figure 8 compares CF reinforced C.P.Al and Al–12Si. Some defects are present in the sample without Si that are not present in the sample containing Si. Alloying Al with 12wt%Si reduces its viscosity by around 13% (see Figure 9) [25]; its reduced viscosity may allow the Al–12Si matrix material to penetrate small pores, resulting in low porosity. Fibre breakage can also be seen in the sample with Si. The addition of Si to Al reduces both solidification shrinkage and coefficient of thermal expansion [26], both of which may reduce fibre breakage during cooling.

AlMMCs have potential applications in automotive and aerospace industries, however ROM predicts that 10 – 15% VF random in-plane CF reinforced Al will increase its stiffness by just 2 – 3GPa. These VFs may be suitable for wear applications, alternatively higher VF preforms will be required for higher strength and stiffness applications. The VF of the preforms used for this work was limited by not being able to use a binder.
Figure 8
SEM images at 2000x magnification of (a) CF/Al–12Si, (b) CF/C.P.AI, porous regions highlighted with white arrows, broken fibres highlighted with a white circle
Figure 9
Effect of alloying additions on the viscosity of aluminium [25]

Figure 4 shows how the degradation of fibres occurs over time at processing temperature. Partial reaction of fibres can be seen after 4 hours (see Figure 4 a), with total degradation of fibres after 8 hours, leaving only wisps of fibres after 24 hours. It is recommended that processing time for CF/Al does not exceed 4 hours, as the reaction with fibres will most likely result in degradation of composite mechanical properties.
Figure 10

SEM images of composite materials after 48 hours at 750°C, (a) CF/Al–12Si at 500x magnification, (b) CF/C.P.Al at 500x magnification, (c) CF/Al–12Si at 2000x magnification, (d) CF/C.P.Al at 2000x magnification

Pelleg et al [11] also fabricated CF/Al–Si composite materials. XRD of their composites showed a small reduction in Al₄C₃ formation, using only 0.5wt%Si. They produced composites using 56MPa (560 bar) of infiltration pressure for 3 minutes, compared with 25 bar for 30 minutes used for this study. The short infiltration time used by Pelleg may have reduced the amount of interfacial carbide formed, by reducing the contact time between the CFs and molten Al alloy.
Although qualitative analysis and XRD did not detect any significant differences between the composite materials fabricated with different amounts of Si in the Al matrix, quantitative testing such as hardness, wear and compressive testing may reveal additional benefits of alloying the Al matrix, and will be considered for future work. Additional work may include mechanical testing of composite materials after heat treatment of varied lengths of time, to determine whether or not degraded fibres will result in a loss of mechanical properties.

6. Conclusions

The following points can be concluded from this work:

(a) CF reinforced Al can be fabricated by GPI, producing composite materials with clean interfaces and low porosity
(b) Heat treatment beyond 8 hours at processing temperature (750°C) results in complete degradation of fibres
(c) Heat treatment or processing time of CF/Al should not exceed 4 hours, as this is when degradation of fibres appears to begin.
(d) Regardless of the amount of Si in the Al matrix, the CFs react with the Al to form $\text{Al}_4\text{C}_3$
7. Appendix

7.1. Sessile Drop Test

7.1.1. Introduction
As molten Al does not wet carbon, fibre coatings have been proposed to solve this issue. The Sessile Drop Test (SDT) has been utilised to determine the optimal CF coating, by dropping molten Al onto coated graphite substrates and measuring the contact angle. If the contact angle can be reduced, this could equate to a reduction in infiltration pressure. Contact angle is directly related to infiltration pressure (see Equation 5):

**Equation 5**

\[ P \propto \cos \theta \]

7.1.2. Set-up
Figure 11 shows the details of the high-temperature drop dispenser. This set-up has been used to remove the oxide layer from the drop. Once the aluminium has melted, the graphite rod is pushed by the spring, forcing the molten aluminium through the \( \Phi 1 \text{mm} \) aperture, which then drops on to the substrate.
A Φ30mm graphite rod was cut into disks 3mm thick and polished with 600-grit (P1200) abrasive paper, for the substrates. The roughness (Ra) of the polished substrates was 0.322μm. A low roughness value is important to contact angle measurement, as surface roughness can widely affect the contact angle [21]. Physical vapour deposition (PVD) was used to coat the substrates with titanium (Ti), titanium nitride (TiN), titanium aluminium nitride (TiAlN) and Al. Electrodeposition was used for nickel (Ni) coatings.

7.1.3. Process

The drop dispenser was set up as shown in Figure 11, and placed in a tube furnace with the aperture directly above a graphite substrate. The furnace was heated to 860°C at a rate of 20°C/min under argon (Ar) flowing at 0.5L/min. Once the Al had melted and dropped onto the substrate, the drop was photographed immediately, and subsequently every 1 minute for 10 minutes. Each substrate material was repeated 5 times.
The contrast of the photos was increased using GIMP software to enhance the drop (see Figure 12). A horizontal line was drawn across the surface of the substrate (base line), and the contact angle between the drop and the substrate was measured by drawing a tangent from the triple point (the point at which the drop meets the substrate) along the edge of the drop (see Figure 13).

![Figure 12](image1.png)

*Figure 12*

*Increased contrast of image to enhance the drop*

![Figure 13](image2.png)

*Figure 13*

*An Al drop with base-line and tangents*
7.1.4. Results and Discussion

Figure 14 summarises the results from the SDT. The contact angle of Al onto graphite can potentially be reduced by 6.25° when coated with Al or TiAlN. From Equation 5, this would reduce the infiltration pressure by just 8.33% (50 bar to 45.84 bar). A contact angle of ≤90° indicates good wetting between 2 materials; the results imply that none of the coatings used would substantially improve wetting between molten Al and C.

Such a small drop in infiltration pressure suggests that either CF coating is not an area to further explore, or the SDT conditions are not representative of the composite infiltration system; the results may be skewed by processing conditions. Although care was taken to ensure that the tube furnace was sealed and flushed with argon before testing began, it is possible that some oxygen remained in the tube, oxidising the Al drop surface and altering the results.

It is unclear at this stage whether the data collected is accurate, so Toho Tenax nickel coated filament yarn will be used as a reinforcement to validate or benchmark the SDT data. Uncoated fibres will be compared with Ni-CFs as a reinforcement, and infiltrated with Al using a range of infiltration pressures, to determine whether Ni-CFs are wetted better by molten Al and therefore can be processed at a lower infiltration temperature.
Figure 14

Contact angle of molten Al-5Si aluminium alloy onto graphite substrates with various coatings
7.2 Directed Carbon Fibre Preforming using Sodium Silicate as a Binder

The main body of this report has highlighted the fact that, without a preform binder, the VF of the composite materials fabricated is low and variable (9 – 15%). A binder was avoided previously in order to simplify the thermodynamic analysis of the system. An alternative preform fabrication method is currently being investigated.

Directed carbon fibre preforming (DCFP) produces randomly orientated discontinuous preforms from a bobbin of CF tow. The CF is chopped and sprayed by a robot-mounted chopper gun in layers; the chopped CFs are held in place on the surface by vacuum [27]. The robot arm can also deposit a binder alongside the fibre, which can be consolidated in situ (see Figure 15).

DCFP has many benefits compared with traditional woven CF preforms. Due to its high level of automation, the manufacturing costs associated with DCFP can be up to 2 orders of magnitude lower than pre-impregnated fabrics [5]. Typically less than 3% of the inputted CF is wasted [28], and a cycle time of 5 minutes is achievable [4].
A liquid sodium silicate (Na$_2$SiO$_3$) binder has been sprayed between DCFP CF layers, using a hand-held spray bottle. Sodium silicate cures within seconds under CO$_2$, so the preform was covered with a vacuum bag and attached to a CO$_2$ canister and a vacuum pump to flush the preform with CO$_2$ and consolidate the preform using the vacuum (see Figure 16).
Figure 16

$CO_2$ curing of sodium silicate binder

Part of the resulting preform can be seen in Figure 17.
The use of CF tows rather than filaments should allow the CFs to pack closer together, resulting in a preform with a higher VF. The VF these types of preforms can be as high as 50% [6]. However, the heterogeneous nature of these preforms results in 44% strength variation [28]. This will need to be investigated further when used as a preform for future metal matrix composite (MMC) fabrication, in addition to; the pressure required to infiltrate a higher VF preform, how the sodium silicate binder affects the wetting behaviour and interfacial reactions, and any other issues as they arise.
7.3 Thesis Plan

Abstract

Introduction

- A background to the project, including information on GHG emissions from transportation, relating this to vehicle mass, and alternative lighter structural materials

Literature Review

- General information about Al, Mg and CFs
- CF/Al and CF/Mg – including mechanical properties, interfacial reactions, fibre coatings
- Fabrication methods – with emphases on GPI
- Environmental aspects of using Mg

Sessile Drop Test and Fibre Coatings

- Background
- Sessile drop test results
- Ni-coated fibres

Preform Fabrication

- Introduction to DCFP or similar process
- Varying amount of sodium silicate binder
  - Preform compaction testing
  - Tensile testing
Gas Pressure Infiltration

- Methodology entailing the details of the GPI process
- Qualitative analysis of composite materials (e.g.: SEM)
- Process parameters affecting quality of composite materials
  - Pressure
  - Temperature
  - Time
  - Alloying
- Quantitative testing
  - Wear
  - Compression

Discussion

This section will bring compare and contrast all results obtained, discuss any anomalies, and put the project into context, discussing its place alongside other similar literature and its potential impact on vehicle manufacture and GHG emissions

Conclusions

Conclusions will be drawn from the entire thesis as a whole, and further work will be proposed
### 7.4 Further Work Plan

<p>| Ni coated fibres | Preform | GP1 at different P | Wear testing | Effect of alloying Mg/Al matrix | Silicon | AZ31 | Other Mg | Wear testing | Compression | Binder study | Binder amount | Compaction | Tension | GPI | Thesis |
|------------------|---------|--------------------|--------------|-----------------------------|---------|------|---------|---------------|-------------|--------------|---------------|------------|----------|------|-------|--------|
|                  |         |                    |              |                             |         |      |         |               |             |              |               |            |          |      |       |        |
| 06.13            | 07.13   | 08.13              | 09.13        | 10.13                       | 11.13   | 12.13| 01.14   | 02.14         | 03.14       | 04.14        | 05.14        | 06.14      | 07.14   | 08.14 |        |</p>
<table>
<thead>
<tr>
<th>Experiment</th>
<th>Aim of experiment</th>
<th>Done so far</th>
<th>Conclusions so far</th>
<th>What needs to be done</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Sessile drop test</strong></td>
<td>To quantify reduction in contact angle between molten Al and CF using CF coatings – reduction in contact angle may result in infiltration pressure reduction</td>
<td>Collected data for:</td>
<td>Coatings can reduce contact angle by 6.25°, equating to 8.33% reduction in infiltration pressure, 50 → 45.84 bar, concluding that either: (a) CF coatings aren't worth the extra processing or (b) Sessile drop data is flawed in some way</td>
<td>Either: (a) Conclude that this avenue is not worth further exploration (b) More reading into sessile drop test to further refine process and repeat (c) Use coated CFs in composite materials to see if drop in infiltration pressure is possible</td>
</tr>
<tr>
<td><strong>Interfacial reactions in C/Al-Si system</strong></td>
<td>To use Si in Al matrix to form SiC at interface rather than Al₄C₃, which is unfavourable</td>
<td>Infiltrated CF mat with:</td>
<td>Well infiltrated samples</td>
<td>SEM on newer composite materials 30-05-13 Depending on results from XRD: (a) If XRD shows no difference between samples, conclude that thermodynamic analysis of system is not correct and write up (b) If XRD shows difference between samples, write up results</td>
</tr>
<tr>
<td><strong>DCFP binder study</strong></td>
<td>To quantify how much sodium silicate binder is required to sufficiently hold together DCFP preform during handling and infiltration To determine if sodium silicate has a positive/deleterious effect on infiltration/mechanical properties</td>
<td>Used sodium silicate with DCFP process Cured binder using CO₂</td>
<td>Process seems to work well Appears to block interface in composite material</td>
<td>Quantify effects of binder using tensile and compressive preform testing Use a DCFPSS preform for GPI and quantify effects of binder using tensile and compressive composite testing</td>
</tr>
<tr>
<td><strong>Ni-coated CFs</strong></td>
<td>To quantify how much (if any) infiltration pressure can be reduced by using Ni-CFs, which should improve the wettability of molten Al on the fibres</td>
<td>Chopped CF tow into ~5mm lengths to use for GPI GPI at 50 bar using Al₁₂Si and both Ni-CFs and uncoated CFs</td>
<td>Infiltration only worked to a limited degree with Ni-CFs, and not at all with un-CFs Possibly due to format of CFs, not in a self-supporting preform</td>
<td>Make fibres into a preform without using a binder Fabricate more composite materials: CF/Al₁₂Si 50 bar 25 bar 5 bar Repeat with: Ni-CF C.P.Ai</td>
</tr>
</tbody>
</table>
8. References


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