Effect of filler metal feed rate and composition on microstructure and mechanical properties of fibre laser welded AA 2024-T3

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ABSTRACT

The influence of aluminium alloy 4043 filler wire feed rate on the weld quality and mechanical properties of high power 5 kW fibre laser welded aluminium alloy 2024-T3 was investigated. Loss of volatile alloying elements such as magnesium and other elements including copper and silicon which all contributed to the hot crack sensitivity was measured using energy dispersive X-ray spectroscopy at different filler wire feed rates. High feed rates of above 4.0 m/min produced instabilities, whereas, low feed rates below 2.0 m/min did not sufficiently modify the chemical composition of the weld pool. The optimum feed rate was found to be in the range between 2 and 3 m/min, where the corresponding dilution ratio of around 9-12% in the weld pool with less than 0.6% silicon content reduced the percentage of Mg₃Si and also decreased the solidification temperature and total shrinkage during freezing. The addition of filler metal reduced the risk of welding defects and improved ductility to over 3.5% and a fairly higher tensile strength of around 380 MPa than without. Microstructural examination showed that the addition of filler wire increased the number of finer dimples within the weld, resulting in a purely ductile fracture behaviour, as well as reduced micro hot cracks and porosities.

KEY WORDS

AA 2024-T3; Fibre laser; Welding; Digital Image Correlation; Mechanical properties; Microstructure

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1 INTRODUCTION

While pure aluminium is not susceptible to hot cracking due to the absence of low melting point eutectic at the grain boundary, the addition of alloying elements such as copper, magnesium and silicon to the 2024 alloy makes its composition crack susceptible. As a result, it is considered to have poor weldability because of its high solute content, resulting in a wide solidification temperature range and increased tendency to form low melting constituents along grain boundary, known as the eutectic composition with lower freezing points than the base metal; and also due to high coefficient of thermal expansion and large solidification shrinkage. Hot cracking is a high temperature cracking mechanism and it can refer to either solidification cracking which occurs intergranularly along the grain boundary of the weld metal during solidification due to the stresses acting on the segregated low melting point constituents [1], or liquation cracking which occurs in the coarse heat affected zone along the fusion boundary as a result of grain boundary liquation of the low melting point constituents.

The weld metal composition in autogenous welding of AA 2024-T3 is identical to the base metal composition, so it is solidification cracking susceptible. Due to the higher equilibrium vapour pressure and lower boiling point of volatile alloying elements such as zinc (917°C) and magnesium (1091°C) compared to aluminium (2519°C), they are excessively and selectively vapourised during laser keyhole welding, and as a consequence, segregation of these elements and dissolution of strengthening precipitates [2] result in degradation of mechanical properties of the weld [3,4]. The vaporisation process involves transport of the alloying element from bulk to surface of the weld pool and then vaporisation at the liquid/vapour interface into the surround gas phases [5]. Moon and Metzbower [6] observed depletion of magnesium in the fusion zone of laser beam welded AA 5456, where a reduction from 5% in the base metal to 4% in the fusion zone occurred. The 20% reduction was considered to be the reason for the reduced mechanical properties of the weld. Alferi et al. [7,8] also observed loss of magnesium of 53% when butt welding 3.2 mm thick AA 2024 at a laser power of 1.6 kW, a welding speed of 10 mm/s and -0.5 mm defocus using disk laser. The reduced Mg content was found to be responsible for keyhole instability which induced macro porosity formation. A part of the reason is because the addition of elements such as zinc, silicon, copper and iron reduce the solubility of hydrogen in liquid aluminium due to strong bonding of Al atoms to these elements, whereas, strong attractive interactions between hydrogen and magnesium, lithium and titanium increase the hydrogen solubility [9]. The 2024 alloy also contains small quantities of silicon which induces the formation of
non-strengthening coarse Mg$_2$Si precipitates, and iron which induces the formation of Cu$_2$FeAl$_7$ phase, both of which reduce the fracture toughness of the weld significantly and deplete the solid solution strengthening Cu solutes in the matrix needed for aged hardening [10].

In order to reduce the cracking susceptibility of AA 2024-T3, it is necessary to adjust the chemical composition and the solidification microstructure of the weld metal to a less susceptible level. Hot cracking sensitivity of the 2024 alloy can be reduced by controlling the solidification process during welding by optimization of welding parameters and avoiding the crack sensitive composition in the weld metal through the use of a suitable filler metal with different chemistry and a lower solidification point than that of the base metal [11]. It is possible to reduce the crack sensitivity by introducing minor eutectic alloy such as Al-Si via the addition of a crack resistant filler metal of the right composition, usually from the 4xxx series alloys. More eutectics are provided by the filler wire to heal the cracks formed during the solidification of the weld pool. In addition, using filler wire also improves the process stability, reduces the tendency to porosity formation and leads to wider volume of the weld pool by compensating for the loss of material due to vaporisation [12]. Improved welding process stability means that macro pores caused by keyhole collapse and instability can be avoided. In addition, the increased silicon content in the molten pool by filler metal addition effectively lowers its hydrogen solubility and therefore, reduces the tendency to micro porosity formation as less hydrogen is dissolved in the molten pool. When filler metal is used, the weld metal composition is determined by the ratio between the amount of the melted base metal and the filler metal in the weld seam. The dilution ratio is usually around 20% for laser beam welding and due to the low fraction of filler metal in the weld, the eutectic alloy 4043 with 5% silicon was used as filler metal in this investigation. The major alloying element of the 4xxx series aluminium-silicon alloys is silicon in amounts up to a maximum of 12% in 4047, with a low melting point suitable as a filler material for welding AA 2024-T3 to eliminate solidification cracking and undercut, without causing brittleness in the resulting welded joint [13]. The 4047 wire however, has lower ductility and is more expensive than the 4043 wire so 4043 was chosen over 4047 for its better ductility. The high percentages of silicon in 4043 filler metal was important for the prevention of solidification cracking and it helped reduce the total shrinkage during freezing. The low solidification temperature and narrow freezing range of around 5°C of the 4043 filler metal ensured that the base metal completely solidifies prior to the weld and reach its maximum strength before become subjected by shrinkage stresses so that the duration in which liquid
metal is affected by this contraction is minimised during cooling. The filler metal was the last part of the weld to solidify. The 4043 filler wire also combined with the base metal to lower the percentage of MgSi$_2$ in the weld and reduce its crack sensitivity. However, the addition of filler metal to the weld also led to a mismatching mechanical properties and reduced the yield and tensile strengths of the weld compared to the base metal. As well as that, in the case where the wire feed rate was too fast and not optimised, it resulted in weld pool instability and formation of welding defects. Therefore, the optimum values for wire feed rate was investigated to produce the best weld quality with reduced crack sensitivity and mechanical performance.

2 EXPERIMENTAL PROCEDURES

Heat treatable aluminium alloy (Al-Cu-Mg) alloy 2024 sheets in the T3 temper condition (i.e. solution heat-treated, cold worked and naturally aged) with 3 mm was used. A 0.6 mm diameter consumable 4043 aluminium filler alloy of a nominal composition Al-5%Si was used when welding 2024 with a filler wire. The chemical compositions of AA 2024-T3 and Aa 4043 are listed in Table 1.

<table>
<thead>
<tr>
<th>Material</th>
<th>Al</th>
<th>Cu</th>
<th>Mg</th>
<th>Mn</th>
<th>Cr</th>
<th>Si</th>
</tr>
</thead>
<tbody>
<tr>
<td>AA 2024-T3</td>
<td>92.1</td>
<td>5.9</td>
<td>1.0</td>
<td>0.6</td>
<td>0.1</td>
<td>0.3</td>
</tr>
<tr>
<td>AA 4043</td>
<td>94.8</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>5.2</td>
</tr>
</tbody>
</table>

A 5 kW continuous wave (CW) ytterbium fibre laser system YLS-5000 from IPG Photonics was used in TEM$_{01}$ mode for laser welding. The beam output of 200 W single-mode active fibres was combined to deliver the beam output to the workpiece via a feed fibre with a diameter of 200 μm. To protect the feed fibre from damage during welding, the feed fibre was coupled into a processing fibre with a diameter of 300 μm which was connected directly to the laser processing head [14], and the beam diameter at focus was 630 μm. The wavelength of fibre laser was 1070 nm, the beam quality factor, M$_2$ was around 7.3, the divergence half angle of the focused beam was 12.5 mrad and the Rayleigh length was about 3.1 mm. The focal length of focusing lens was 300 mm, and the diameter of the focusing lens was 50 mm. The focal length of the collimator lens was 100 mm and the diameter of the collimator lens was 50 mm. A beam parameter product (BPP) of less than 2.5 mm mrad was formed.

The filler wire feed rate was the only parameter which was changed while all the other parameters were kept constant. An independent investigation was conducted to optimise the remaining parameters such as laser power, welding speed and focal position. All specimens were welded using
a laser power of 4.9 kW, a welding speed of 3.0 m/min, +4 mm defocus and helium gas shielding as shown in Figure 1.

![Image](image.png)

**Figure 1 Full penetration bead on plate welding of AA 2024-T3 using AA 4043 filler metal at different feed rates**

The effect of varying the filler wire feed rate was studied as shown in Table 2. A BINZEL Master-Feeder system was used to supply filler metal into the leading edge of the weld pool, ahead of the laser beam impingement point at an angle of 45 or 60° with the workpiece. Industrial grade helium with 99.999% purity was used at a flow rate of 15-20 l/min. The shielding gas was supplied to protect both top and underside of the weld. The coaxial shielding gas was delivered via the weld nozzle to protect molten pool and the back protecting shield gas was supplied via the shielding gas path in the copper insert to protect back weld. Both top and bottom surfaces of each specimen were brushed with a stainless steel wire brush and then cleaned using an industrial grade unbuffered 99.9% pure acetone before welding.

<table>
<thead>
<tr>
<th>Laser power (kW)</th>
<th>Speed (m/min)</th>
<th>Focal position (mm)</th>
<th>Wire feed rate (m/min)</th>
<th>Dilution ratio (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>0.0</td>
<td>0.0</td>
</tr>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>1.5</td>
<td>6.3</td>
</tr>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>2.0</td>
<td>8.5</td>
</tr>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>2.5</td>
<td>10.3</td>
</tr>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>3.0</td>
<td>12.3</td>
</tr>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>4.0</td>
<td>16.2</td>
</tr>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>5.0</td>
<td>20.1</td>
</tr>
<tr>
<td>4.9</td>
<td>3.0</td>
<td>+4.0</td>
<td>7.0</td>
<td>27.4</td>
</tr>
</tbody>
</table>

The microstructural constituents of the weld are revealed by using suitable chemical etchants. Keller’s reagent which is a mixture of 95% distilled water, 2.5% HNO₃, 1.5% HCl and 1.0% HF was used to etch AA 2024-T3 weld specimens by immersing for 10-30 seconds fresh. Macroscopic and microscopic inspection of transverse sections of metallographic welded specimens followed the test procedures specified in BS EN 1321 [15]. Microscopic examination with a magnification within 50 to
500 times also with etching to reveal features of welded joints. An optical microscope (OM), Zeiss Axio Scope A1 was used for microstructural examination. Energy dispersive X-ray spectroscopy (EDX) in an environmental scanning electron microscope (SEM), Hitachi S-3400N VPSEM was used to determine chemical compositions of the specimens at an accelerating voltage of 15 kV, an emission current of 76 μA, a working distance of 6.8 mm, an elevation of 35° and a live time of 50 seconds. Scanning electron microscope (SEM) was also used to examine the fracture surface of selected tensile specimens after tensile testing in the base metal and the welded joints at various magnifications to determine the fracture behaviour and the presence of welding defects.

The quality of welds produced in this experiment were evaluated against a set of welding acceptance criteria from several international standards on welding as shown in Table 3:

<table>
<thead>
<tr>
<th>Standard</th>
<th>Level</th>
<th>Face width (mm)</th>
<th>Root width (mm)</th>
<th>Porosity (mm)</th>
<th>Undercut (mm)</th>
<th>Underfill (mm)</th>
<th>Reinforcement (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AWS D17.1</td>
<td>Class A</td>
<td>N/A</td>
<td>N/A</td>
<td>≤ 0.99</td>
<td>≤ 0.05</td>
<td>≤ 0.13</td>
<td>≤ 0.99</td>
</tr>
<tr>
<td>BS EN ISO 13919-2</td>
<td>stringent B</td>
<td>N/A</td>
<td>N/A</td>
<td>≤ 0.90</td>
<td>≤ 0.15</td>
<td>≤ 0.15</td>
<td>≤ 0.65</td>
</tr>
<tr>
<td>BS EN 4678</td>
<td>AA</td>
<td>≤ 4.00</td>
<td>≤ 2.50</td>
<td>≤ 0.90</td>
<td>≤ 0.15</td>
<td>≤ 0.30</td>
<td>≤ 0.55</td>
</tr>
</tbody>
</table>

The main criteria assessed were the top and bottom weld widths, the ratio of root to face width, $R_w$, the depth of undercut and underfill, the size of weld porosity and the height of reinforcement or excess weld penetration. Any weld showing a lack of penetration or crack were rejected according to AWS D17.1, BS EN ISO 13919-2 and BS EN 4678. Other welding imperfections such as overlap and spatter were also identified. The $R_w$ was used to evaluate the processing stability of full penetration welding and a value of 0.6 determined by Chen et al. [19].

The influence of filler wire feed rate on microstructural transformations and variations of local hardness profiles of AA 2024-T3 welds was evaluated by measuring hardness in the direction perpendicular to the transverse weld cross-sections along three lines located 0.5 mm away from the top and the bottom surfaces, with a 1 mm gap between them. Micro-hardness measurements were performed on weld cross-sections perpendicular to the weld line using a Zwick Roell Z2.5 (ZHU 0.2) hardness testing machine at a load of 100 g and a dwell period of 15 seconds at a speed of 60 μms⁻¹ to characterise the whole hardness profile across the weld seams up to the base metal. To avoid work hardening contributions from the adjacent indents, indentations were separated by 200 μm and the lines of indentation were separated by 500 μm.
Uniaxial tensile tests were conducted at ambient temperature on a 200 kN Instron electromechanical universal testing machine in ram displacement control at a constant crosshead speed of 1 mm/min. Tensile properties were determined from two butt-welded 3 mm thick AA 2024-T3 tensile specimens one welded with and the other without filler metal. Digital image correlation (DIC) technique was used to capture digital images of the deforming surface of the specimens during uniaxial tensile deformation every second to compute the corresponding displacement and strain. Images were captured using the 2D DIC technique at the minimum focus distance of 0.45 m and at a resolution of 3456 × 2304 pixels. The DIC area of analysis was selected manually to process the images only within the specimen, with a resolution of 310 × 1800 pixels. Additionally, a facet size of 20 pixels and a facet step of 16 pixels, giving an overlap of 4 pixels, were set up according to the configuration with the average speckle diameter and the resolution of the image in order to ensure that the results are accurate.

Table 4 lists the welding parameters of the specimens tested. In order to evaluate quantitatively the differences in mechanical properties between the autogenously welded specimen and the specimen welded using an optimum feed rate of 2.6 m/min as found in this investigation, the development of strain distribution during loading was determined.

<table>
<thead>
<tr>
<th>Welding mode</th>
<th>Laser power (kW)</th>
<th>Welding speed (m/min)</th>
<th>Focal position (mm)</th>
<th>Wire feed rate (m/min)</th>
<th>Dilution ratio (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Autogenous</td>
<td>2.9</td>
<td>1.5</td>
<td>+4</td>
<td>-</td>
<td>0.0</td>
</tr>
<tr>
<td>Filler</td>
<td>2.9</td>
<td>1.5</td>
<td>+4</td>
<td>2.6</td>
<td>10.5</td>
</tr>
</tbody>
</table>

3 RESULTS AND DISCUSSIONS

Microstructure of the weld with and without filler metal was examined at 50x and 500x magnifications as shown in Figure 2. The specimen which was welded autogenously contained large solidification cracks on the weld top surface, whereas, the one welded in the presence of filler metal displayed no visible crack within the weld. It was found that in both cases the centre of the FZ showed the formation of characteristic equiaxed dendrites, and columnar dendrites near the FZ boundary. Epitaxial continuous growth of columnar grains was observed in the direction of thermal gradients in the FZ, with the same crystallographic orientation to that at the FZ line. Dendritic growth was reported by Watkins et al. [20] for alloys containing less than 5% weight copper, in this case AA 2024, with the dendrites being α-Al and with either CuAl2 precipitates or CuAl2-Al eutectic as the inter-dendritic.
phase. The FZ of AA 2024 mainly consisted of α-Al phase with a surrounding eutectic CuMgAl$_2$ phase as it contained magnesium.

During welding of AA 2024, the low melting point eutectic with a wide range of freezing temperatures segregated in the grain boundaries and formed the low melting point constituents, which were rejected by the solidifying columnar grains. The amount of eutectic liquid between grains were large enough to form a thin, continuous grain boundary film during solidification at a depressed liquidus and solidus temperatures because of magnesium, compared to the bulk solidus temperature. The solidus temperature was further suppressed due to a lack of diffusion resulting from rapid non-equilibrium solidification during welding. The shrinkage strains was proportional to the coherence range between the first formation of mushy stage by dendrite interlocking and the solidus, so a wider coherence range increased the tendency for solidification cracking [21]. When the amount of liquid available during the freezing process was insufficient to fill in the spaces between the solidifying grains at the centre, then micro-cracks such as those observed in Figure 2 were formed due to the lack of material and high shrinkage strains in the weld pool [22]. Equiaxed dendritic structure on the other hand, reduced the crack sensitivity due to the abundance of liquid metal between grains which were able to
deform more easily under stresses [23], and the lower coherent temperature range resulting from the formation of equiaxed dendrites at a later stage in freezing [24]. In addition, the fine isotropic grain structure of equiaxed grains unlike coarse anisotropic columnar grains, increased the resistance to crack formation and propagation [25] by distributing the low melting point segregates over a larger grain boundary area and also relieved local shrinkage strains developed during freezing more efficiently [24]. Equiaxed grain formation is important for the grain refinement of welds but due to the high solidification rate and thermal gradient, it is often considered difficult to obtain. Instead, columnar grain growth is favoured and there is a small chance of equiaxed grain formation, resulting in predominantly coarse, low ductility columnar grain structure in the FZ [26]. The use of 4043 filler metal which has a freezing range of around 5°C enabled rapid solidification of welds and reduced the time for shrinkage during solidification, and therefore, micro-cracks were not observed in specimens welded with filler wire.

The weld face and root dimensions were initially measured as shown in Figure 3. It was found that the top and the bottom weld width of the specimens increased with increasing filler wire feed rate. The change in the bottom width with feed rate was relatively small but quite large for the top width. The top weld width of specimens measured at all feed rates passed the criterion in BS EN 4678, of 4.0 mm, whereas the bottom weld width was above the maximum of 2.5 mm at all feed rates and therefore, failed the criterion. However, as the weld quality in general was good in terms of welding defects and Rw, and also because there were no weld quality acceptance criteria related to weld width specified in AWS D17.1 and BS EN ISO 13919-2, it was not possible to judge the specimens using the criteria on face and root weld widths in BS EN 4678 alone but also had to consider other factors as well. It was found that the magnesium content in the weld measured by EDX decreased with increasing feed rate from around 0.80% at 0 m/min to 0.65% at 7.0 m/min, which obviously increased the silicon content as well, from around 0.10% at 0 m/min to 0.88% at 7.0 m/min, illustrated in more details in Figure 5. The Rw was above 0.6 at all feed rates so the processing stability for full penetration welding was high. A trend was observed where the Rw decreased with increasing feed rate, meaning that as mentioned above, the rate of change in the top width was greater than the bottom width with the feed rate. Underfill was observed when autogenous welding, above the maximum limit of 0.13 mm in AWS D17.1 and 0.15 in BS EN ISO 13919-2 but below 0.30 mm in BS EN 4678. The depth of underfill was reduced by more than half at 1.5 m/min which passed all criteria so welding with filler wire reduced
the formation of underfill defects and even eliminated at higher feed rates. As expected the height of reinforcement or excess weld metal increased with increasing feed rate and at 7.0 m/min, it was above the most stringent limit of 0.55 mm in BS EN 4678 but below that in BS EN ISO 13919-2 and AWS D17.1, of 0.65 and 0.99 mm respectively. For the rest of the specimens, it was less than 0.55 mm. The depth of undercut was below the maximum limit of 0.15 mm in BS EN ISO 13919-2 and BS EN 4678 at all feed rates, but was above the 0.05 mm in AWS D17.1 at the two highest feed rates of 5.0 and 7.0 m/min. The problems with welding defects including surface porosity, reinforcement and undercut were found to be the most significant at 5.0 m/min and 7.0 m/min. It was possible that the feed rate was too high at these feed rates which supplied too much filler metal to the weld pool for the given laser power and welding speed. High feed rates produced instabilities, whereas, low feed rates did not sufficiently modify the chemical composition of the weld pool [11].

The weld shape was similar at all feed rates with an hourglass shape but with larger weld widths with increasing feed rate. The small underfill observed at 0 m/min was reduced at 1.5 m/min and completely removed at 2.0 m/min. The weld quality was good between 2.0 and 4.0 m/min but at 5.0 m/min, clustered pores with a maximum diameter of 0.21 mm. The criterion in AWS D17.1 for surface porosity specifies at least 8 times the size of larger adjacent imperfection and a smaller maximum pore size of 0.75 mm compared to 0.99 mm for subsurface pores. BS EN 4678 and BS EN ISO 13919-2 on the other hand, specify the same diameter of 0.99 mm for surface pores but with the distance between the individual pores in clustered porosity greater than ¼ of the material thickness. Although the size of surface pores was small, they were too close to each other so were unacceptable. These surface pores were not observed at 7.0 m/min but instead large undercut

![Figure 3](image_url)
defects on the top surface and excessive penetration on the bottom surface was found. Therefore, it was concluded that the best weld quality in terms of morphology was produced when the filler metal feed rate was in the optimum range of 1.5 to 4.0 m/min.

![Figure 4](#) Transverse sections of welds and weld top bead profiles produced with different filler metal feed rate at a laser power of 4.9 kW, a welding speed of 3.0 m/min, +4 mm defocus and with helium shielding gas

The effect of silicon addition on the crack sensitivity of AA 2024-T3 is shown in Figure 5. The weight content of magnesium and silicon in the weld was measured using energy dispersive X-ray spectroscopy on specimens welded with different filler metal feed rates. It was found that the magnesium level dropped whereas, the silicon level increased with increasing feed rate. As it can be seen from the crack sensitivity curves for aluminium in Figure 5, the crack sensitivity is the maximum when the Cu content is approximately 3%, Si is 1%, and Mg is 1.5%. AA 2024-T3 contained approximately 4.5% Cu which may initially have indicated that it has relatively low crack sensitivity. However, it also contained a small amount of Mg close to the critical level of 1.5% in the base metal, which increased the crack sensitivity by widening the coherence range, and depressing the solidus temperature but not the highest temperature of coherence [27]. Segregation of the low boiling point alloying elements such as Zn and Mg caused hot cracking at the grain boundaries due to the shrinkage strains during the solidification process. The presence of Si as well as Mg in the AA 2024-T3 base metal increased the risk of inducing coarse Mg2Si precipitates so the maximum content of Si in the 2024 alloy was required to be less than 0.7% [28]. According to Davis [29], the sensitivity
decreases rapidly if the Si content exceeds 1.5%. The dilution of the weld pool with excess silicon by welding with the 4043 filler metal effectively reduced the percentage of Mg2Si in the weld by combining with the base metal. Also, the addition of silicon to the weld lowered the solidification temperature and decreased the total shrinkage during freezing as mentioned previously to prevent cracking. As a result, the peak of the solidification crack sensitivity curve for Al-Mg and Al-Mg2Si shifted away from the crack sensitive ranges. The silicon content in the welded specimens were detected to be less than the recommended 0.6% to avoid the crack sensitive range of Al-Si up to the feed rate of 3.0 m/min but above 0.6% at higher feed rates of 4.0, 5.0 and 7.0 m/min. Therefore, the solidification crack sensitivity was minimised by welding at a filler wire feed rate of 2.0 to 3.0 m/min with a dilution ratio of 8.5 to 12.3%.

![Graph showing weight percentage (% weight) of main alloying elements in the weld as a function of filler metal feed rate obtained using energy dispersive X-ray spectroscopy (EDX) and aluminium crack sensitivity curves showing the effects of different alloy additions (Figure 5 b) modified from [21,30)]](image)

**Figure 5 a)** Weight percentage (%) of main alloying elements in the weld as a function of filler metal feed rate obtained using energy dispersive X-ray spectroscopy (EDX) and **b)** aluminium crack sensitivity curves showing the effects of different alloy additions (Figure 5 b) modified from [21,30])

### 3.1 Micro-hardness

Micro-hardness testing in the transverse weld bead cross-sections as illustrated in Figure 6 showed that all AA 2024-T3 welds were under-matched with the lowest hardness in the FZ. The hardness in the BM was the highest as expected. It was also found that the hardness in the HAZ was greater than in the FZ but lower than in the BM. Micro-hardness increased as a function of the distance from the weld centre in which the FZ had a hardness of around 90-100 HV, the HAZ hardness in the range of 100-120 HV and the BM hardness in the range of 130-140 HV. The HAZ adjacent to the FZ showed hardness values close to that in the FZ whereas, the HAZ adjacent to the BM showed a hardness close that in the BM. Since the extent of the FZ and the HAZ was very small, the resulting hardness gradient was very steep. On the other hand, the hardness distribution was relatively uniform across the FZ in most specimens.
The dissolution or loss of strengthening precipitates and alloying elements, softening in the FZ, and over-aging in the HAZ were the main causes of hardness degradation during welding process. The effect of grain growth with respect to strength was of minor importance but instead mainly influenced by modification of precipitates [31]. Softening in the FZ was caused by microstructural changes as a result of very high temperatures experienced in the FZ and the associated rapid heating and cooling rates during welding. The heating action of the laser led to segregation of the strengthening elements, magnesium and copper, and their hybrids (intermetallic compounds), formation and growth of non-strengthening coarse precipitates, dissolution of strengthening precipitates and uniform re-distribution of precipitating elements during heating which then froze due to fast cooling rates [32]. In addition, softening can also be attributed to violent vaporization of low boiling point magnesium and element variation resulting from the filler dilution [33] as observed in Figure 5 a) showing lower Cu and Mg contents with increasing filler metal feed rate. The hardening effect was therefore, removed and the mechanical properties of the weld degraded. The hardness in the FZ was similar to the hardness measured in a fully solution treated and quenched AA 2024 of around 80 HV [34]. Even though the FZ partially recovered its hardness by natural ageing at room temperature for several days after welding, the effect was small due to inhomogeneous distribution of solute atoms. Loss of volatile elements such as magnesium and zinc for strengthening also contributed to lowering the hardness in the FZ by affecting the weld pool chemistry. The welding thermal cycle also affected the precipitation behaviour in the HAZ such as dissolution, precipitation and coarsening so the HAZ was divided into two different microstructural regions of partially melted zone and over-aged zone. The hardness in the partially melted zone decreased due to dissolution of strengthening precipitates during melting and segregation of alloys during solidification. In the over-aged zone, coarsening of the strengthening semi-coherent S' phase as well as transformation to the non-strengthening incoherent stable S phase reduced the hardness [35]. As it can be seen from Figure 6, changing the filler metal feed rate does not significantly affect the heat input and therefore, its effect on micro-hardness was relatively small but rather influenced the weld width, where increasing the feed rate increased both the face and the root weld widths.
Figure 6 Micro-indentation hardness distributions of fibre laser welded AA 2024-T3 welds as a function of filler wire feed rate
3.2 Global and local tensile properties

Figure 7 shows the full field longitudinal strain distribution at different percentage of the fracture time at 25, 50, 75 and 99%. The strain profiles across the weld showed a sharp strain gradient from the weld centreline to the BM at the onset of final fracture. The maximum strain localisation was located in the FZ for both specimens, whereas, the minimum occurred in the BM. The strain distribution was uniform and symmetrical about the weld centreline throughout the deformation with relatively larger local strains of around 11% in the FZ for both specimens welded with or without filler metal.

![Figure 7](image)

**Figure 7** Full field longitudinal strain distributions in the loading direction for fibre laser welded AA 2024-T3 showing the development of strain localisation relative to the time to fracture

The strain measured in the BM was of the order of only 0.5-1.0% at failure which indicated that it was still under elastic loading. Although the boundaries of different characteristic regions corresponding to the BM, the HAZ and the FZ cannot be easily identified from the strain maps, their locations were measured and marked outside the processed regions prior to testing and also it was possible to determine their extent by examining the highly non-uniform strain distribution across the weld at different load levels. It was obvious from these strain maps that the stiffness of the welded specimen was a result of the stiffness of the three different microstructural regions

The global tensile behaviour of the welded specimens and the BM and the corresponding mechanical properties are shown in Figure 8. It was possible to create any size and number of gauge lengths on the processed DIC images and so the global stress and strain curves were determined for a 25 mm gauge length equal to that of the extensometer used. For mechanical characterisation of the welded
joints, the elastic modulus, yield strength, ultimate tensile strength and elongation to failure were determined as listed in Table 5.

Table 5 Results of tensile properties determined from tensile testing welded joints

<table>
<thead>
<tr>
<th>Welding mode</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>Elongation (%)</th>
<th>E (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Filler</td>
<td>370</td>
<td>277</td>
<td>3.7</td>
<td>63.6</td>
</tr>
<tr>
<td>Autogenous</td>
<td>364</td>
<td>283</td>
<td>2.4</td>
<td>66.6</td>
</tr>
<tr>
<td>Unwelded</td>
<td>463</td>
<td>358</td>
<td>14.9</td>
<td>69.1</td>
</tr>
</tbody>
</table>

Global tensile test results showed significant losses in ductility and tensile strength in the welded specimens compared to the unwelded BM due to plastic strain localisation and increased constraint within the lower strength weld region of the welded joint for the composite gauge length. Only moderate variations in the yield strength and elastic modulus were observed, while considerable differences in elongation to failure and ultimate tensile strength were measured. The addition of filler metal reduced the risk of welding defects and resulted in a significantly greater ductility over 3.5% and a fairly higher tensile strength of around 380 MPa than the other specimen.

Figure 8 Global stress and strain curves obtained from tensile testing welded specimens

The influence of local material behaviours in the various weld zones on the overall weld response was determined using the DIC by assuming an iso-stress condition for all specimens, where the global stress was considered as the corresponding local stress at any point within the analysed displacement data field. The local strain data were derived from the DIC by measuring the local strain over each individual region, which were then plotted against the global stress data to obtain the local tensile properties in the FZ and the HAZ on either side of the FZ (HAZ1, HAZ2) as shown in Figure 9. As it was only possible to obtain the full tensile response in the weakest region where strain localisation occurred, the stress and strain curves in the stronger regions such as the BM were not obtained. The high hardening rate
and low ductility observed in the global stress and strain curves of the welded specimens proved that strain localisation occurred in the weaker regions of the weld. In fact, as all the AA 2024-T3 specimens were under-matched, the local strain evolution in the FZ or the HAZ was successfully calculated up to complete fracture, which was not possible to determine from standard tensile specimens without the DIC because of the narrow size of the FZ. It was found that the longitudinal strains measured in the weld was much higher than the measured global fracture strains which indicated that the strain distribution within the 25 mm gauge length was not uniform but highly localised in the weaker FZ or HAZ so the overall behaviour was dominated by that of the weakest component of the specimen and minimal plastic deformation occurred outside the weld [36]. The maximum strain was reached in the FZ for both specimens.

P=2.9 kW, V=1.5 m/min, f=+4 mm, filler metal feed rate=2.6 m/min (10.5% dilution ratio)

Figure 9 Local mechanical responses in FZ and HAZ constructed from full field DIC tensile tests compared to overall responses

Figure 10 shows the relationship between longitudinal strain distributions across the weld at the onset of final fracture obtained using the DIC and micro-hardness distributions in and around the respective welds from micro-indentation hardness testing. The strain values in the BM were small tending to zero with increasing distance from centre of weld, whereas, the strain increased in the HAZ at the same positions where the hardness decreased so the hardness measurements were used to confirmed the extent of the HAZ as well as the FZ where the hardness was the minimum with only small variations. The strain distribution was more uniform in the specimen welded with filler wire, with a wider weld width.
Figure 10 Local strain distributions across the welded joint obtained using the DIC at the onset of final fracture superimposed on micro-indentation hardness distributions measured in the transverse weld cross-sections

3.3 Microstructural observations using SEM

Scanning electron microscopy (SEM) analysis was performed on the fracture surfaces of the AA 2024-T3 DIC tensile specimens as shown in Figure 11 to examine the fracture behaviour and the existence of welding defects such as hot cracks and porosities. It was observed that for both specimens inspected, failure occurred within the weld, either in the FZ or in the HAZ/FZ boundary but not in the BM due to weld under-match.

For the specimen which was welded without filler wire, a rough and irregular fracture surface, consisting of microvoids and dimples that act as microvoid nucleation sites, was observed. This indicated that the specimen failed in a more ductile manner where fracture initiated by microvoid coalescence and then by dimple rupture. Rough fracture surfaces with fine equiaxed dimples are the characteristic features of ductile transgranular fracture mode. Hydrogen induced spherical shaped pores of different sizes ranging from around 100 to 300 μm with round tip dendrites, were identified on the fracture surface in clusters which were responsible for crack initiation during deformation. Porosities reduced the effective cross-sectional area of the welded joints and therefore, caused stress concentrations which deteriorated the strength of the joints in proportion to the reduction of the cross-sectional area.

For the specimen which was welded with filler wire at an optimum rate of 2.6 m/min (10.5% dilution ratio), even greater amount of finer dimples was found to dominate the fracture surfaces which is the characteristic feature of purely ductile fracture. A significant amount of localised microscopic weld plasticity was observed which improved the tensile strength and ductility of the respective welds as previously determined from tensile testing. While intergranular inter-dendritic micro hot cracks and
porosities were detected in the other specimen, they were almost or completely absent in this specimen so a significant improvement in tensile strength and ductility were obtained. In fact, porosity level less than 3% of the total volume does not affect much the yield and tensile strength of a material but may affect ductility [37].

P=2.9 kW, V=1.5 m/min, f=+4 mm, filler metal feed rate=2.6 m/min (10.5% dilution ratio)

![SEM fracture morphology of DIC tensile test specimens at 500x magnification showing different modes of failure and presence of welding defects](image)

4 CONCLUSIONS

Welding with filler metal reduced the crack sensitivity of AA 2024-T3 but it was also important to optimise the filler metal feed rate to avoid the formation of welding defects and keyhole instability. High feed rates produced instabilities, whereas, low feed rates did not sufficiently modify the chemical composition of the weld pool. The optimum feed rate was found to be around 2-3 m/min (8.5-12.3% dilution ratio) which minimised the crack sensitive range of Al-Mg, Al-Mg2Si and Al-Si according to the measurements obtained by EDX.

Changing the filler metal feed rate does not significantly affect the heat input and therefore, its effect on micro-hardness was relatively small but rather influenced the weld width, where increasing the feed rate increased both the face and the root weld widths.
Only small variations in the yield strength and elastic modulus were observed, while considerable differences in elongation to failure and ultimate tensile strength were measured. The addition of filler metal reduced the risk of welding defects and improved ductility to over 3.5% and a fairly higher tensile strength of around 380 MPa than without. The strain distribution was more uniform in the specimen welded with filler wire, with a wider weld width.

Microstructural examination showed that the addition of filler wire increases the number of finer dimples within the weld, resulting in a purely ductile fracture behaviour. A significant amount of localised microscopic weld plasticity was observed which improved the tensile strength and ductility.

5 ACKNOWLEDGEMENT

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