Optimisation of process parameters and weld shape of high power Yb-fibre laser welded 2024-T3 aluminium alloy

J. Ahn, L. Chen, E. He, J.P. Dear, C.M. Davies

1. Introduction

Laser beam welded aluminium structures are increasingly being used in the aircraft industry for their lightweight high performance. Aluminium alloy 2024-T3 has excellent tensile to yield strength ratio at elevated temperatures as well as high ductility, fatigue resistance and fracture toughness. Yet, it also possesses number critical inherent characteristics such as high thermal conductivity and thermal expansion coefficient (large solidification shrinkage), high reflectivity to the laser, low boiling point elements, high hydrogen solubility in liquid state and aluminium oxide films, which make fusion welding difficult. Some of the alloying elements such as copper, magnesium and silicon make its chemical composition sensitive to hot cracking by widening the solidification temperature range and promoting the formation of low freezing point eutectic constituents along grain boundaries. Hot cracking can involve intergranular solidification cracking along grain boundaries of the weld metal during solidification as a result of stresses acting on the segregated low melting point constituents, and liquation cracking in the coarse heat affected zone along fusion boundary due to grain boundary liquation of low melting point constituents [1].

As the 2024 alloy has a range of freezing temperatures due to the presence of alloying elements, the solidification strains are proportional to the coherence range between the formation of coherent interlocking dendrites and the solidus temperature; and mechanical properties of the alloy in the temperature range of crack formation. The effective solidus temperatures are depressed by rapid non-equilibrium solidification during welding. The areas with higher freezing points solidify first from the fusion boundary towards the weld centreline. The low melting point constituents are rejected by the formation of a coherent interlocking dendrite solid network and the remaining liquid along grain boundaries form continuous thin inter-dendritic liquation films which persist down to low temperatures. A crack is formed when the liquid films cannot withstand the thermal shrinkage strains developed in the solidifying weld metal and the supply of residual eutectic liquid available to fill the incipient cracks caused by the shrinkage strains between grains during freezing is insufficient [2].

The hot cracking sensitivity of the 2024 alloy can be reduced by controlling the solidification process during welding through optimisation of welding parameters and avoiding the crack sensitive composition in the weld metal by using a suitable filler metal with different chemistry and a lower solidification point than that of the parent material [3]. Ahn et al. [4] investigated the influence of aluminium alloy...
4043 filler wire feed rate on the weld quality and material properties of fibre laser welded AA 2024-T3 and identified the optimum feed rate with a dilution ratio of around 9–12% in the weld pool with less than 0.6% Si content which reduced the fraction of Mg$_2$Si and decreased the solidification temperature and total shrinkage during freezing. Chen et al. [5–7] examined how the choice of shielding gas including argon and helium affected the welding defect formation and microstructure of fibre laser welded AA 2024-T3 and concluded that better weld quality was achieved using helium at lower welding speeds due to reduced plume effect as helium has a higher ionisation potential than argon and therefore, minimised porosity formation. However, the use of argon shifted the magnesium content in the weld further away from the critical level for crack sensitivity than helium under identical welding conditions.

Jones et al. [8] and Sutton et al. [9] studied friction stir welding (FSW) of AA 2024-T351 as an alternative solution to fusion welding to avoid hot crack issues, where the formation of brittle solidification products and grain boundary liquation cracking were minimised because it is a solid-state welding process. On the other hand, fibre laser has received an increasing interest for welding based on the advantages it offers over existing lasers such as better laser efficiency, higher output power, superior beam quality, longer working distances, lower maintenance, reduced cooling requirements, smaller footprint and more compact design [10]. Ytterbium laser has a wavelength of 1.07 μm which is similar to that of the Nd:YAG laser, so it has the same advantages of shorter wavelength lasers such as fibre optic beam delivery, improved weldability and lower minimum power density threshold for keyhole mode welding compared with the CO$_2$ laser [11].

The 2xxx and 7xxx series aluminium alloys are the two main series used for aircraft structures. The 2xxx series Al-Cu-Mg alloys such as the AA 2024-T3 are primarily used in tension dominated structures. Although the 7000 series alloys have higher strength than the 2000 series alloys, they have lower damage tolerance (DT) and stress corrosion resistance so their use is restricted to compression limited structures. The 6xxx series alloys, particularly in any stress environment. There are two types of 7xxx series welding processes, so it had very limited use for welding applications especially in any stress environment. There are two types of 7xxx series alloys, the Al-Zn-Mg alloy which can easily be welded and the Al-Zn-Mg-Cu alloy which is crack sensitive like the 2024 laser. Laser welding of the weldable 7xxx series alloys has been investigated by numerous researchers such as Liu et al. [13], Weston et al. [14] and Squillace et al. [15], Zhang et al. [16,17], Allen et al. [18], Enz et al. [19], and Paleocrasas and Tu [20]. On the other hand, very few investigations on laser welding of difficult to weld materials, in this case, AA 2024-T3 can be found in the literature.

In a lot of industrial applications especially when welding relatively easy to weld lightweight materials, the selection of processing parameters is almost always based on previous experience of the operator. Such decision is often made without a clear fundamental scientific understanding of the intricate relationship between welding parameters and metallurgical properties of welded joints mainly because reasonable weld quality can still be achieved without knowing it. Investigating this complex relationship is very time consuming and expensive in an industrial perspective. As a result, a great deal of previous research on welding applications and even still, base its conclusions on a very limited number of welding parameters, in many cases chosen without a strong scientific evidence. However, in this case, as AA 2024-T3 is crack sensitive and prone to welding defects, the usual way of processing based on intuition and experience should not be used as it does not lead to the prevention of critical and unpredictable failure of welded AA 2024-T3 aircraft components. Therefore, it is crucial to fully examine the properties and performance of fibre laser welded AA 2024-T3, optimise the welding procedure and processing parameters based on fundamental understanding of the material and process relationship so that high-quality welds without welding defects and with good mechanical properties can be produced consistently [21]. Hence, this investigation attempted to determine the influence of processing parameters including laser power, power density, welding speed and beam focal position on weld morphology, the formation of defects and microstructure of AA2024-T3 welded joints.

2. Materials and experimental procedures

2.1. Materials

3 mm thick sheets of 2024 aluminium alloy in the solution heat-treated, cold worked and naturally aged T3 temper were used. The sheets were cladded with high-purity aluminium surface layers to increase corrosion resistance Table 1 lists the chemical compositions of the 2024 alloy and that of the 0.6 mm diameter consumable 4043 aluminium filler alloy.

2.2. Laser welding process

A 5 kW IPG YLS-5000 fibre laser was used to bead on plate weld with the configuration as shown in Fig. 1. The focal diameter of the laser was 450 μm, the beam quality factor (M$^2$) was around 8.25, the divergence half angle of the focused beam was 12.5 m rad and the Rayleigh length for the multimode beam, scaled with M$^2$ was around 18 mm. A beam parameter product (BPP) was around 2.81 mm mrad.

Root shielding was done by purging with 99.999% argon before welding and then supplying at a flow rate of 101/min through the bottom copper insert during welding. Face shielding was done coaxially via weld nozzle at 20 l/min to shroud over the molten weld pool during welding. An air-knife as shown in Fig. 1 was used to supply the gas from the side, perpendicular to the optical axis to protect the lens system from spatters. 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° to the workpiece in dragging configuration. As illustrated in Fig. 1, the processing head was tilted at 5° to prevent laser beam back reflection of the laser beam from damaging the optics and re-entering the laser cavity. A visual inspection of both top and bottom surfaces was done before welding to identify any source of surface contaminations including moisture and oil (sources of hydrogen), dust and surface oxides on the parent material, filler wire and from the atmosphere. Rigorous cleaning of the specimens involved degreasing to remove oil, dust and loose particles using an industrial grade unbuffered 99.9% pure acetone and then brushing with a stainless-steel wire brush (mechanical cleaning) to remove surface oxide and contaminates immediately before welding to reduce the risk of weld porosity.

A wide range of processing parameters as listed in Table 2 was utilised. The range of laser power used was from 1.9 to 4.9 kW, where the maximum level used was close to the limit of the welding system and any level lower than the chosen minimum failed to weld. The laser power was directly obtained from the laser power indicator of the fibre laser welding system and the welding speed was controlled by the robot. The focal distance and focal spot diameter were accurately determined using PROMETEC’s laser scope UFF100 type beam mass analyser and PROLAS software. The range of welding speed used was

![Table 1 Chemical composition of AA 2024-T3 (Wt. %).](image-url)
from 1.1 to 6.0 m/min, where the maximum level was chosen based on process stability and the minimum level was based on economic feasibility of using fibre laser for welding aircraft components as processing time is important. The range of focal position used was from +4 to −4 mm, where 0 mm corresponds to the beam focused on the top surface of the workpiece and the beam diameter increased with increasing magnitude of defocusing.

### 3. Results and discussion

It was possible to produce high-quality fibre laser welded 2024-T3 joints such as the one shown in Fig. 2a) via optimisation of process parameters, where a laser power of 4.9 kW, a welding speed of 3.0 m/min, a focal position of +2.0 mm above the top surface of the welds which were incompletely penetrated through-thickness or contained macro-cracks were rejected without a tolerance margin. Other minor defects such as overlap, and spatter were identified but not considered fully when assessing the weld quality. \( R_w \) was used to evaluate the dimensional effect of the weld seam geometry on process stability for full penetration, where a value of 0.6 as previously found by Chen et al. [25] was used.

### 2.3. Metallographic specimen preparation

Two transverse weld cross-sections were extracted from the weld by electrical discharge machining (EDM) for each set of welding parameters. The specimens were ground, polished and then chemically etched using Keller’s reagent by dipping for 30s. The compositions of the etchant were 95% distilled water, 2.5% HNO₃, 1.5% HCl and 1.0% HF. While this duration was sufficient for the parent metal, depending on the specimen, a longer etching time was required to reveal the weld microstructure. The etched specimens were examined under optical microscope (OM) and scanning electron microscope (SEM), and energy dispersive X-ray spectroscopy (EDS) was used to determine the chemical compositions 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 s.

### 2.4. Welding quality acceptance criteria

Welding and inspections were performed in accordance with the procedures and requirements specified in AWS D17.1, BS EN ISO 15614-11 and BS EN ISO 13919-2 [22–24]. These standards provided guidance on levels of imperfections in laser beam welded joints in metallic materials. Visual and metallographic examinations were performed to assess the weld quality using the criteria listed in Table 3. The main criteria assessed were the face (top) and root (bottom) weld widths, the ratio of root to face width (\( R_w \)), the size of undercut, underfill, porosity and reinforcement or excess weld penetration. Four repeated measurements of the weld seam geometry were taken on two separately cut specimens for each welding parameter set to identify the measurement uncertainty. According to the criteria listed in Table 3, the welds which were incompletely penetrated through-thickness or contained macro-cracks were rejected without a tolerance margin.

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### Table 2

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### Table 3

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workpiece, a 1.0 mm diameter 4043 consumable filler metal at a feed rate of 5.1 m/min and argon gas shielding at a flow rate of 20 l/min were used. The weld exhibited an hourglass-shaped transverse cross-section and consisted of fusion zone (FZ), heat affected zone (HAZ) and parent material (PM). The equilibrium precipitate phases of the PM were dominantly semi-coherent CuMgAl2 with a small proportion of CuAl2, both a function of the weight ratio of Cu and Mg. The dark precipitates observed in Fig. 2b) were intermetallic compounds, varying in size, shape and chemical composition, formed during solution heat treatment and natural ageing in T3 temper before welding. The HAZ as shown in Fig. 2c) and d) was the boundary region in between the PM and the FZ, which experienced peak temperatures below the melting point of the PM but high enough to exceed the solvus curves, which affected its microstructure. It consisted of two parts, the region near the FZ where the dissolution of the strengthening precipitates occurred and the region next to the PM where over-ageing by coarsening of the strengthening precipitates, semi-coherent CuMgAl2 phase, and then a transformation to the incoherent stable CuMgAl2 phase occurred [26,27]. The overaged zone had the maximum amount of the coarser and sparser stable phase CuMgAl2. As such phase is known to not respond well to post-weld ageing [26,27] without prior solution heat treatment after welding, post weld heat treatment (PWHT) for the 2024 alloy was not recommended. The width of the HAZ was less than 0.1 mm, with only a few grains across, due to the high thermal gradient, low heat input and fast cooling rate, which were characteristics of fibre laser beam welding so only limited metallurgical modifications were made.

The FZ consisted mainly of α-Al phase with surrounding eutectic CuMgAl2 phase. The initial solidification occurred epitaxially near the outer FZ boundary with a fine grain zone of planar front growth from the PM as shown in Fig. 2. The planar front solidification switched to dendritic grain growth of elongated columnar dendrites towards the centre due to constitutional supercooling. The dissolution of precipitates in the FZ resulted in its softening as measured by the microhardness distribution from the corresponding weld transverse cross-section as shown in Fig. 3, where the hardness value in the FZ dropped considerably to around 90 HV compared to that in the PM of around 140 HV and progressively increased in the HAZ to around 110 HV. The effect was more pronounced in the FZ than the HAZ due to higher temperatures experienced and so a greater dissolution of strengthening phases occurred. The fast welding speed and high thermal gradients at the FZ boundary led to an elongated weld pool which promoted the formation of columnar dendritic structures. As it can be seen from Fig. 2e), the centre of the FZ showed the formation of characteristic equiaxed dendrites, and from Fig. 2f) columnar dendrites near the FZ boundary. The epitaxial continuous growth of columnar dendrites was observed in the direction of thermal gradients in the FZ, with the same crystallographic orientation to that at the FZ line. The size of the dendrite cells was influenced by the heat input, laser power and welding speed, all of which were responsible for the temperature gradient and the solidification rate. Decreasing the welding speed led to larger
dendrite cells in the FZ, wider FZ width and larger grain size in the HAZ. Dendritic growth is known to occur for alloys which contain less than 5% weight copper and so does AA 2024, with the dendrites being α-Al and with either CuAl₂ precipitates or CuAl₃-Al eutectic as the inter-dendritic phase [28].

3.1. Effect of power density on weld shape

The effect of power density on microstructure, weldability and morphology was investigated by comparing three different combinations of laser power and welding speed, which generated similar nominal heat inputs including: (i) 2.9 kW and 1.8 m/min, (ii) 3.9 kW and 2.4 m/min, and (iii) 4.9 kW and 3.0 m/min. The heat input (or line energy) associated with the power density used was determined by dividing the laser power by the welding speed. The laser power and the welding speeds were chosen to produce nominal heat inputs within a close range between 96.5 and 98.5 J/mm.

It was necessary to maintain the power density above 10⁸ W/cm² for keyhole mode welding, where the material becomes vaporised before dissipating all the heat by conduction, below 10⁸ W/cm², where the laser energy does not penetrate the workpiece [29]. At the same time, it had to be kept below 10⁷ W/cm² to minimise defects such as undercut, underfill, excessive penetration and spatter.

Increasing the power density allowed reduction of weld width by enabling the use of a faster welding speed with higher thermal efficiency. Higher power density expanded the operating window but also led to more laser-induced plume and loss of alloying elements by vaporisation, thus, encouraging the formation of welding defects. Visual inspection of the top and bottom weld surface for power densities of 2.05, 2.76 and 3.47 MW/cm² as illustrated in Fig. 4, showed no visible critical macro defects such as cracks or surface pores. All three seams were fully penetrated.

Both the top and bottom weld widths as shown in Fig. 5 were less than the maximum acceptable face and root widths of 4.0 and 2.5 mm, respectively, according to BS EN 4678. The processing stability for full penetration welding (Rw) with a value above 0.6 was satisfied at all power densities with values of around 1.2. The maximum height of reinforcement found at 2.76 MW/cm², of around 0.46 mm, was below the limit and acceptable. Figs. 5 and 6 also show that the undercut depth was the largest at the 3.47 MW/cm², of around 0.22 mm, above the lower limit of 0.05 mm and the upper limit of 0.15 mm as listed in Table 3. An excessive undercutting was produced due to the increased amount of molten parent material lost through evaporation and spatter so the gap during solidification was insufﬁciently backﬁlled. The size of undercuts observed at 2.05 and 2.76 MW/cm² were both less than 0.15 mm but greater than 0.05 mm. However, such limit was considered very conservative and therefore, the measured undercuts were acceptable as they still met the requirement of the less stringent criterion.

The weld seams as illustrated in Fig. 6, were all hourglass shaped, a typical shape that is found in a stable full penetration laser weld. An undercut at 3.47 MW/cm² was observed on both surfaces, whereas, it was only found on the top surface at 2.05 MW/cm². Only a small undercut was observed on the top surface at 2.76 MW/cm² but, a relatively large excessive root penetration was observed.

Despite the acceptable visual appearance and absence of porosity, intergranular micro-cracks smaller than 0.5 mm were observed as shown in Fig. 7. It indicates that fracture occurred at the grain boundaries and the fractured surface was dendritic with blunt crack tips. As the amount of liquid available during the freezing process was insufficient to fill in the spaces between the solidifying grains at the centre, cracks formed due to the lack of material and high shrinkage strains acting on the low melting point eutectics in the weld pool. The weld seam was unable to withstand the contraction stress during solidiﬁcation because the molten metal in the weld pool could not ﬁll in the gaps due to inadequate supply or narrow channels between solidifying grains.

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 dendrites. The amount of eutectic liquid between grains was large enough to form a thin, continuous grain boundary ﬁlm during solidiﬁcation at a depressed liquidus and solidus temperatures compared to the bulk solidus temperature. The solidus temperature was further suppressed due to a lack of diffusion resulting from rapid nonequilibrium solidiﬁcation during welding. The shrinkage strains were proportional to the coherence range between the first formation of the mushy stage by dendrite interlocking and the solidus, so a wider coherence range increased the tendency for solidiﬁcation cracking.

Optical microscopic examination of the FZ at 500x magnification identiﬁed an increasing area of equiaxed dendritic structures with decreasing power density, which effectively reduced the solidiﬁcation rate and the thermal gradients in the weld to promote the growth of equiaxed dendrites. Equiaxed dendrites, unlike coarse anisotropic columnar dendrites, increased the resistance to crack formation and propagation by distributing the low melting point segregates over a larger grain boundary area and relieved local shrinkage strains developed during freezing more efﬁciently [33]. Equiaxed dendrite formation is important for the grain reﬁnement of welds but due to the high solidiﬁcation rate and thermal gradient. Fine equiaxed dendritic structure with sufﬁcient liquid between grains can increase solidiﬁcation crack resistance, by decreasing the coherent temperature range thus allowing it to deform more easily than columnar dendritic structure. CET was promoted by the reduction of solidiﬁcation rate and thermal gradient in the weld pool.

The loss of alloying elements including magnesium with a low boiling point of 1091 °C and high equilibrium vapour pressure was measured using an area scan at the weld centreline via energy dispersive spectrometry (EDS) analysis in the transverse cross-section of the welds. The vapourisation of these elements occurred at the surface of the weld pool into the surrounding gas phases. Table 4 shows the analysed chemical composition of the parent material compared to welds of three different power densities obtained from the EDS spectrum. An average loss in magnesium content of around 0.5%, from 1.2 to 0.7% was observed in the welds compared to the parent material (PM), but only a small difference was found with respect to the power density. The reduced Mg content in the weld was responsible for keyhole instability which increased the risk of porosity formation and hot crack susceptibility. A change in power density was expected to affect the extent of alloying elements losses by influencing the temperature of the molten metal in the weld pool. However, as the nominal heat inputs used among these specimens were similar, the difference was small. Nevertheless, the rate of vapourisation and the volume of weld pool can be controlled by varying welding parameters such as welding speed and laser power. The extent of Mg vapourisation was a function of the welding speed. As the welding speed increased, there was less time to...
evaporate Mg so high speed reduced elements losses. The vaporisation of the alloying elements was lower when using faster welding speed due to shorter laser and material interaction time, so the welding process was more stable. One of the reasons for the increased fraction of aluminium in the weld metal compared to the parent material was because the Alclad layer melted and diluted the weld metal as well as from contributions of a lower fraction of alloying elements.

3.2. Effect of laser power on weld shape

Laser power controls the incident laser power density so the effect of changing laser power on weld microstructure, weldability and morphology were examined using three different sets of parameters. The weld surface profiles in Fig. 8 indicated that the weld widths increased with increasing laser power due to greater power density and line energy (assuming a constant speed) supplied to the workpiece. A poor weld seam appearance including solidification crack observed at high laser powers of 3.9 and 4.9 kW, with a power density above 2.0 MW/cm², was due overheating, where the total heat input was greater than 2.0 J/mm given a welding speed of 2.0 m/min. A lower total heat input of less than 1.6 J/mm at the same laser powers, given a welding speed of 3.0 m/min, as shown in Fig. 8 b) resulted in more inconsistent, thinner, elongated and weak seams but produced a few underfill defects both on top and bottom weld surfaces at 3.9 kW. Incomplete penetration was observed on the bottom surface at 1.9 kW given a welding speed of 3.0 m/min, where either or both the power density of less than 1.2 MW/cm² and total heat input of less than 1.0 J/mm were insufficient.

A linear relationship between weld width and laser power was found as shown in Fig. 9. The top weld width at all laser powers was below the limit of 4.0 mm, whereas, the bottom weld width was close to or greater than the limit of 2.5 mm specified in Table 3 at above 3.9 kW. The Rw was greater than 0.6 which is required to produce a stable full penetration weld, except where full penetration was not achieved at 1.9 kW. In the case where full penetration was achieved at 1.9 kW, a large underfill with a depth of 0.4 mm was formed instead as shown in Fig. 10, which was significantly greater than both the lower limit of 0.13 mm and the upper limit of 0.15 mm as listed in Table 3. At laser powers above 2.9 kW, the top and bottom widths became similar, as indicated by the Rw values approaching close to one. This trend can also be clearly observed in Fig. 11, where the weld shape at 1.9 kW was V-shaped, which at 2.9 kW was nearly rectangular, and hourglass shaped.

![Fig. 5. Influence of power density on a) top and bottom weld widths and b) weld width ratio, undercut and reinforcement.](image_url)

![Fig. 6. Transverse sections and weld top bead profiles produced using three different power densities (weld transverse cross-sections extracted from the welds in Fig. 4).](image_url)
At 3.9 and 4.9 kW with a wider bottom.

At higher laser powers, evaporation and expulsion of weld metal resulted in undercuts at the toe of the parent material and reinforcement at the weld centre. However, the height of reinforcement measured was smaller than the limit at all laser powers, so the laser power had a negligible effect on excess weld metal. According to Fig. 10, both undercut and underfill were considerable at 1.9 kW, underfill was greater than the upper limit at 2.9 kW given 3.0 m/min and undercut was over the upper limit given 2.0 m/min and +4 mm. Undercut was noticeable at 3.9 kW for all three sets, with a depth greater than the lower limit of 0.05 mm. It was caused by the low viscosity of the molten weld pool which could be improved by the addition of filler metal or optimising face and root shielding.

Underfill at 1.9 kW, given 2.0 m/min was clearly visible and the weld quality was unacceptable as shown in Fig. 11. As mentioned above, the weld shape changed from a V to hourglass shape with increasing laser power. The weld quality at 2.9 kW without defocusing was good because only small reinforcement and undercut and no underfill were produced. Root concavity was observed at 3.9 kW but still below the limit. The weld seam at 4.9 kW was slightly asymmetrical with an unacceptable size of the undercut on one side of the weld toe. Therefore, a moderate laser power of 2.9 kW provided the optimum welding condition given a welding speed of 2.0 m/min.

Table 4

<table>
<thead>
<tr>
<th>Power Density (MW/cm²)</th>
<th>Al (Wt %)</th>
<th>Ca (Wt %)</th>
<th>Mg (Wt %)</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.05</td>
<td>85.65</td>
<td>6.06</td>
<td>0.70</td>
</tr>
<tr>
<td>2.76</td>
<td>86.19</td>
<td>5.97</td>
<td>0.69</td>
</tr>
<tr>
<td>3.47</td>
<td>87.57</td>
<td>5.84</td>
<td>0.63</td>
</tr>
<tr>
<td>Parent</td>
<td>84.52</td>
<td>6.05</td>
<td>1.21</td>
</tr>
</tbody>
</table>

at 3.9 and 4.9 kW with a wider bottom.

Fig. 7. Equiaxed dendrite structure at the weld centreline and columnar dendrites on the sides at 50x, and detailed images of solidification micro-cracks at 500× magnifications for three different power densities.

Fig. 8. Variations in face and root surface quality of weld seams with laser powers given a) v = 2.0 m/min and f = 0 mm, b) v = 3.0 m/min and f = 0 mm, and c) v = 2.0 m/min and f = +4 mm.
A root micro-pore with a diameter of 0.22 mm was found on the lower half of the weld at 1.9 kW as well as a surface pore on the top weld seam as illustrated in Fig. 11b), with a size much less than the limit of 0.90 mm. Increasing the heat input while failing to achieve full penetration resulted in enhanced vaporisation of alloying elements which induced keyhole instability and gas occlusions [30,31]. The likely source of micro-porosity with a diameter of less than 0.20 mm [32], was hydrogen dissolved in the weld pool, considering its small spherical shape. Entrapment of gas bubbles occurred as the solidification rate was too fast given the welding speed of 3.0 m/min and the rejected hydrogen at the solid-liquid interface, which exceeded the low solid solubility limit, failed to escape the solidifying weld. This is because the solubility of hydrogen is very high in the liquid phase, about 20 times greater than in solid but drops significantly during cooling in the solid state. Another less likely mechanism for the formation of spherical pores was the inclusion of shielding gas, which is more frequently observed in pulsed wave laser beam welding [33]. The key method of reducing porosity was to remove hydrogen sources before and during welding or to produce a hydrogen oversaturated weld by increasing the solidification rate, for example, by increasing the welding speed. Therefore, the weld quality was acceptable above or equal to 2.9 kW given a welding speed of 3.0 m/min.

Increasing the laser power from 3.9 to 4.9 kW led to more material being melted so the contraction strain became larger and caused larger cracks to form. Such behaviour was worsened by defocusing the laser beam to + 4 mm as shown in Fig. 11c), which increased the beam diameter and the area affected. Also, a slower solidification rate due to higher laser power caused impurity elements to diffuse into the molten pool and the resulting weakened microstructure of the weld in the middle was susceptible to cracking [34]. As it can be seen from Fig. 11c), cracks were in the middle of the weld seam. It was determined that for the given welding speed of 2.9 m/min and + 4 mm defocus, the overall weld quality was unacceptable for all laser powers and therefore, required alternative parametric studies such as varying welding speed to improve the weld quality.

### 3.3. Effect of welding speed on weld shape

Welding speed is an important processing parameter which controls the total heat input, thermal gradient and heating and cooling rate. Its effect on weldability, microstructure and morphology was studied using three different parameter sets. As welding at speeds above 2.0 m/min given a low laser power of 1.9 kW, resulted in incomplete penetration and formation of pores and underfill, the possibility of using slower welding speed (increased the total heat input per unit length) was examined to resolve these issues.

Reducing the welding speed all the way down to 1.0 m/min as indicated by the bottom weld seams in Fig. 12a) failed to produce full penetration. A discontinuity in the top weld seam at 1.3 m/min was an anomaly caused by a sudden fault in the welding system and was ignored in the analysis. While reducing the welding speed increased the total heat input from 1.0 to 1.9 J/mm, the power density of 0.7 MW/cm² at 1.9 kW was found to be still insufficient to produce full penetration, meaning the minimum threshold power density was not achieved. Increasing the welding speed decreased the weld width as shown in Fig. 13 due to reduced heat input, resulting in the less molten material. The determined power density at 1.9 kW of 0.672 MW/cm² at the focus and even less with a + 4 mm defocus, was below the 1 MW/cm² required to form a keyhole. The depth to width ratio at 1.9 kW was close to one, commonly observed in conduction mode.

At 2.9 kW, with a greater power density of 1.1 MW/cm², full penetration was achieved at all welding speeds examined from 1.1 to 2.0 m/min. as shown in Fig. 12b). In addition, no visible surface macro-sized defect was observed and the amount of spatter around weld seams was lower. The observed widening of the weld width for lower welding speed suggested more heat conduction perpendicular to the welded joint with decreasing welding speed. Both the top and bottom weld widths below 1.5 m/min were excessive, being larger than the thickness of the welded sheet, can have detrimental effects on the weld strength. They were below the limit of 1.7 m/min as shown in Fig. 13. The Rw at 1.7 m/min was less than 0.6 and the root width compared to the face...
width was considerably narrower, so full penetration stability was low. Reinforcement was found as excessive root penetration rather than excess weld metal on the top surface but still smaller than the limit of 0.55 mm as shown in Fig. 14. An underfill 0.48 mm deep was measured at 2.0 m/min and failed all acceptance criteria. The fast cooling rate at the given speed resulted in not enough time for the weld to solidify with a sufficient liquid metal flow to fill the weld seam. It can act as a crack initiation point and reduce the cross-sectional thickness and have a negative impact on the weld mechanical properties. At slower welding speeds where the solidification rate was lower, the weld seam was sufficiently backfilled. The depth of undercut at 1.5 and 1.7 m/min was greater than of the lower limit of 0.05 mm but less than the upper limit of 0.15 mm. This meant that even at 2.9 kW, the welding speed available for full penetration welding was limited to a relatively slow welding speed range lower 3.0 m/min, which in terms of manufacturing productivity in the aircraft industry is not ideal so the possibility of using faster welding speeds at higher laser power was studied. As expected at 2.0 m/min, given a high laser power of 4.9 kW with a power density of 3.1 MW/cm² and total heat input of 2.5 J/mm, centreline solidification cracking was observed as shown in Fig. 12c) due to the excessive heat input in the molten weld pool. Full penetration was achieved at all welding speeds from 2.0 to 6.0 m/min. No critical defect was observed by visual inspection at faster welding speeds. It was found that increasing the welding speed reduced cracking susceptibility through grain refinement and formation of a finer dendrite structure. The dendrite arm spacing decreased with increasing welding speed, so

![Fig. 11. Variations in transverse cross-section and top bead profile of weld seams with laser powers given a) v = 2.0 m/min and f = 0 mm, b) v = 3.0 m/min and f = 0 mm, and c) v = 2.0 m/min and f = +4 mm.](image-url)
the weld metal microstructure became finer at higher welding speeds (faster cooling rate). On the other hand, hot cracking was still a possibility at faster welding speeds as the resulting increased solidification rate may induce thermal shrinkage strains and increase stress gradient, resulting in a high crack initiation rate and reducing the time for residual liquid along grain boundaries to back-fill the initiated cracks. This meant that a sufficiently low welding speed together with a relatively low heat input was required to minimise the risk of transverse cracking that is caused by an elongated temperature distribution in the welding direction. Wider welds were found to be less crack sensitive than narrow welds so a lower welding speed, on average produced higher quality welds provided a sufficiently high-power density. Increasing the welding speed also improved the keyhole stability as shown in Fig. 13. The Ra was above 0.6 and very consistent for all welding speeds studied. However, the top and bottom widths were only acceptable above 2.0 m/min due to excessive bottom width. Reinforcement was measured at all welding speeds but was below the limit and showed no clear correlation between the height of reinforcement and welding speed as indicated in Fig. 14. An underfill was only found at 2.0 m/min, whereas, undercut was observed between 3.0 m/min and 6.0 m/min.

A large 0.59 mm diameter macropore was found at 1.0 m/min given a laser power of 1.9 kW as shown in Fig. 15a), which lied within the range between 0.3 and 0.6 mm for classifying macropores. Macropores, unlike micropores, are formed due to keyhole collapse during welding because of large differences in melting and boiling points of the parent material and the alloying elements in this case, magnesium in AA 2024. The vaporisation of Mg during welding increased the tendency to form macropores by affecting the keyhole pressure, which was affected by the welding speed. Partial penetration welds were as those in Fig. 15a) were found to be more prone to porosity than full penetration welds because the path for the escape of gas bubbles was only available via the top surface, whereas, in full penetration welds, the pores which formed at the root of the weld could also escape from the bottom surface rather than travelling up to the top surface. Porosity was not observed at higher welding speeds as expected, either because the keyhole stability increased with welding speed or there was not enough time for nucleation of porosity.

The weld shape at 2.9 kW was influenced by the welding speed as illustrated in Fig. 15b), in such a way that a slower welding speed resulted in a considerable increase in the weld width due to the increased heat input. The difference between the top and bottom weld widths became greater with increasing welding speed, where the weld shape changed from a rectangular shape with flat edges to a deep narrow V shape with a single weld centreline boundary. The top weld seam profile was relatively flat up to 1.7 m/min and then a large underfill was found at 2.0 m/min. A flat top surface geometry proved that a good weld quality was produced as there was no weak point acting as a source of stress concentration in the weld seam. Although relatively wide weld width close to the limit was produced at 1.5 m/min

Fig. 12. Variations in the face and root surface quality of weld seams with welding speed given a) 1.9 kW and +4 mm defocus, b) 2.9 kW and +4 mm defocus and c) 4.9 kW and +4 mm defocus.

Fig. 13. Variations in the face and root weld widths and weld width ratio with welding speed.
compared to those at 1.7 and 2.0 m/min, still the weld quality was good with no underfill and acceptable undercut. Therefore, the best weld quality was obtained at the welding speed of 1.5 m/min. The weld centre of the specimens consisted of equiaxed dendrites and fine columnar dendrites close to the fusion boundary. The number of equiaxed dendrites decreased with increasing welding speed. A faster welding speed promoted a fine dendritic structure and grain sizes resulting from low heat input. The columnar grains growth was epitaxial at faster welding speeds. In contrast, at slower welding speeds, the columnar grains curved away from the normal to the welding direction and instead, aligned parallel to the welding direction.

An undercut was formed as a groove in the parent material along the edges of the weld as shown in Fig. 15c) at faster welding speeds given a laser power of 4.9 kW, because during welding the molten liquid metal was drawn along the edges of the weld bead by surface tension and piled up along the weld centre and prevented from wetting.
back during the rapid solidification. On the other hand, neither undercut nor underfill defects were detected at 4.0 and 5.0 m/min, where the top weld geometry appeared to be flatter. However, a small micropore related to hydrogen in the weld pool was located close to the weld centreline with a diameter of 0.064 mm at 5.0 m/min. It was determined that its influence on the weld quality and performance was negligible according to the three acceptance criteria. By increasing the welding speed further to 6.0 m/min, hydrogen porosity was effectively removed because there was not enough time for nucleation of hydrogen pores due to rapid cooling and solidification rate at higher welding speed. As the welding speed increased, the critical hydrogen concentration required to form was greater [35]. The critical welding speed at which the formation and growth of hydrogen porosity were prevented was 5.0 m/min. As the welding speed increased, the weld width decreased because the heat input was small due to less exposure of the laser beam on the workpiece at higher welding speeds, so the depth to width ratio at 4.0 and 5.0 m/min was large and the FZ size was minimised. The weld shape at slow welding speeds was hourglass shaped but as the welding speed increased, the shape gradually became straighter with flat edges especially above 4.0 m/min. At 4.0 m/min, the weld quality was optimised with the minimum number of welding imperfections.

3.4. Effect of focal position on weld shape

Defocusing was achieved by varying the vertical distance of the laser beam relative to the top surface of the workpiece. It had the effect of changing the incident beam size and power density as well as the resultant weld bead geometry, and therefore, carefully selected to optimise the weld quality. Variations in specific energy caused by changing the focal position led to different material responses during welding [30]. The power density was the greatest when the beam is focused and gradually drops with increasing axial distance from the focal point. Defocusing widened the weld pool and keyhole size, which in some cases improved the processing stability by reducing the excessive power density at the focus that causes excessive vaporisation [36]. On the other hand, it increased the amount of reflected laser energy from the surface of the workpiece and resulted in fluctuations in the energy absorbed.

The effect of increasing the magnitude of defocus was examined using a laser power of 3.9 kW and a welding speed of 2.0 m/min as excessive undercutting was observed when using a focused beam. A considerable variation in the surface appearance of weld face and root with defocusing was observed as shown in Fig. 16b). It was clear that poor weld seam characteristics including a crater and rough surface were observed with a focused beam. Defocusing improved the weld seam appearances with a more uniform weld width and shape along the welding direction with fewer spatter and crater. In addition, the effect of reducing the magnitude of defocus was examined using a laser power of 4.9 kW and welding speed of 3.0 m/min as excessive undercutting was observed when using a focused beam. A considerable variation in the surface appearance of weld face and root with defocusing was observed as shown in Fig. 16b) when checked by visual inspection. As the chosen set of parameters exhibited hot cracking issues, the interaction between defocusing and filler metal addition was examined by adding AA 4043 filler metal to eliminate such defect. At either zero or negative focal positions, the laser-material interaction took place above the focal point where the power density and beam energy absorption were lower. As a result, the filler metal melted not by direct contact with a laser beam but instead via heating within the weld pool. Due to both fast welding speed and filler metal feed rate, there was insufficient interaction time to fully melt the filler wire before entering the weld pool. This caused the filler metal to reflect part of the laser beam energy, meaning that the focal position was not optimised with respect to the wire feed when positioned below or on the surface of the workpiece. This could be avoided by positive defocusing so that the filler wire is incident with the laser beam impingement point. The weld surface appearance improved significantly with a smoother surface as shown in Fig. 16c) compared to the autogenous welds in Fig. 16b). Both the top and bottom surfaces displayed a concavely shaped weld seam without crater formation.

A wider bottom weld width compared to the top one observed at negative focal positions at a laser power of 3.9 kW and welding speed of 2.0 m/min, was due to the lower power density threshold for keyhole formation at negative focal positions where greater irradiance and more efficient laser-material interaction occurred within the weld pool [35]. Also, beam loss due to reflection was reduced by multiple reflections at the keyhole wall and effectively increased the vapour pressure at the bottom of the keyhole and the beam absorptivity [37]. Such enhanced beam concentration effect at the lower half of the weld was proven by the $R_w$ values being close to one at + 4 mm and −2 mm as illustrated in Fig. 17. While the top weld width at all focal positions were within the limiting width of 4.0 mm, the bottom weld width was above the limit of 2.5 mm at −4, −2 and +4 mm. Undercut defect as observed at 0 mm could be avoided defocusing but caused undercuts to form instead. Due to relatively high total heat input of 2.1 J/mm for the given set of parameters, the depth of undercut was greater than the upper limit of 0.15 mm at focal positions of −2 and +2 mm, and smaller than the upper limit −4 and +4 mm but greater than the lower limit of 0.05 mm as shown in Fig. 18. Reinforcement at all focal positions was acceptable and below the lower limit of 0.55 mm. Therefore, the weld quality was found to be acceptable at −4, −2 and 0 mm but unacceptable at +2 mm due to an excessive undercut and at +4 mm due to surface crack as shown in Fig. 19.

No obvious increase in bottom weld width with negative defocusing was observed when using a laser power of 4.9 kW and a welding speed of 3.0 m/min. With a total heat input of 1.6 J/mm that is 0.4 J/mm lower than when using a laser power of 3.9 kW and a welding speed of 2.0 m/min, both the top and bottom weld widths were smaller than the limit at all focal positions. In addition, the calculated $R_w$ values did not show much change with defocusing, with values ranging between 0.8 and 1.0, which were greater than the reference value of 0.6 required for full penetration stability. Reinforcement was observed at all focal positions but significantly less than the limit. No underfill was observed at all focal positions. No undercut was formed at −2 and 0 mm, formed but below the lower limit at −4 mm, below the upper limit at +2 mm but above at +4 mm. The weld quality was good at −4, −2 and 0 mm, acceptable at +2 mm but unacceptable at +4 mm due to excessive undercut formed.

On the other hand, when welding with filler wire, the bottom width considerably increased with defocusing as shown in Fig. 17c). The bottom weld width increased at positive focal positions because of the enhanced interaction between laser and filler wire where the fraction of reflected laser energy was reduced and the absorption of the beam energy (power density) was maximised. Such mechanism had less effect on the top weld width. It was less than the limit at all focal positions. The bottom weld width, on the other hand, was greater than the limit at +2 and +4 mm but can be acceptable when considering the error margin. The $R_w$ above 0.6 at all focal positions so stable full penetration was achieved. The transverse weld cross-section of the weld at the focus was symmetrical with similar top and bottom weld widths. A decrease in $R_w$ with increasing focal position indicated a greater difference between the top and bottom widths. A large undercut observed at +4 mm when autogenous welded, was avoided by welding with filler wire and reached a size close to the most stringent criteria of 0.05 mm in AWS D17.1 as shown in Fig. 18c). No undercut was observed between +0 mm to −4 mm. Reinforcement became an issue at +0 mm due to the high filler wire feed rate, reaching almost as high as 0.5 mm. However, it was easily removed by defocusing.
The weld shape at +0 mm as illustrated in Fig. 19a) was flat with straight fusion boundaries. It increasingly became hourglass shaped with curved fusion boundaries with increasing focal position. Weld centreline cracking was observed at +4 mm was due to excessive top weld width caused by increased amount of parent material being melted and lost through evaporation. This meant that the gap formed during welding was insufficiently backfilled to heal the initiated crack. The reduced power density also contributed to increasing solidification crack susceptibility.

Fig. 19b) shows a micro-crack that is less than 0.5 mm at +4 mm, at the weld centreline. Such crack was unidentifiable by visual inspection and only by examining under an optical microscope at a high magnification. Also, due to the size, location and quasi-two-dimensional geometry of micro-cracks, it is difficult to detect by using NDT methods due to limited interaction. The weld shape at all focal positions was hourglass shaped with curved fusion boundaries and relatively flat top weld surfaces. Excessive penetration was observed at +0 mm but still retained a good weld quality according to the acceptance criteria. The ideal weld profile was obtained at −2 mm but was also acceptable at other focal positions except at +4 mm, where large undercut and root...
shrinkage groove were found. This showed that it is possible to produce high-quality welds even when using high laser power and welding speed, over a range of focal positions, with characteristics and qualities suitable for aerospace applications.

Welding with filler metal improved the weld shape as shown in Fig. 19c) especially at 0 and +4 mm. An hourglass-shaped weld cross-section without defect was observed at all focal positions. However, an overlap was identified at +2 mm at the weld toe (or root) caused by molten metal flowing on to the surface of the parent material without fusing to it. The cause of overlap is often a large weld pool size but can also be the inaccurate positioning of the workpiece or presence of contaminants such as adherent surface oxides. An asymmetrical weld cross-section at +2 mm indicated a possible misalignment of the wire feed position from the plane defined by the laser beam optical axis and the welding direction. Even a slight misalignment of 0.4 mm when welding with filler wire transverse to the welding direction can result in an asymmetrical weld. This can be fixed by reducing the weld pool size, accurate positioning and alignment of the wire feed, laser beam and the workpiece, and more importantly, careful cleaning of the workpiece prior to welding.

4. Future opportunities

While the weldability of AA 2024-T3 fibre laser welded joints was systematically studied in this investigation, deterioration of mechanical properties of welded joints and the resulting load-carrying capacity, especially tensile strength due to the softening behaviour in AA 2024-T3 [38–40] and hydrogen and keyhole induced porosity formation caused by dissolution and evaporation of strengthening elements such as Mg [41] still need to be fully assessed through extensive experimental research. Examination and testing of welds can be conducted in accordance with acceptance level C and D of BS EN ISO 15614-11:2002 standard for electron and laser beam welding [42] including micro-hardness test, tensile tests, metallographic and radiographic examination, for example, using synchrotron radiation analysis to image welding defects such as porosity and crack as well as to measure residual stresses.

Better service performance of laser welded joints can be obtained by reducing the porosity level as discussed in this investigation as well as via optimisation of other processing parameters [5–7]. Attaining higher tensile properties or less softening effect will require minimising crack formation in the weld by shifting the chemical composition away from the crack sensitive range via filler metal addition with an optimised feed rate. Post weld artificial ageing heat treatment to recover the strength to the same level as the parent material can be problematic as it can lead to over-ageing in the HAZ.
5. Conclusions

A novel approach for welding crack and porosity sensitive aluminum alloy 2024 by using advanced high-power fibre laser was developed. The weld quality, shape and microstructure of 2024 bead on plate welds were optimised by determining the ideal combination of process parameters. Solidification cracking along grain boundaries of the weld metal due to segregation of low melting point constituents was reduced by carefully controlling the solidification process during welding and the weld shape as it dictated the solidification pattern via parametric optimisation. A sufficiently wide weld pool reduced the risk of solidification cracking and shrinkage strain during solidification, whereas, an excessive weld width resulted in the crack formation due to insufficient weld pool backfill and high shrinkage strain.

Increasing the power density expanded the operating window but led to more laser-induced plume and loss of alloying elements by vaporisation, thus, encouraging the formation of welding defects. Decreasing the power density, on the other hand, increased the resistance to crack formation and propagation by promoting the formation of the equiaxed dendritic structure instead of columnar dendrites. Micro-cracks of less than 0.5 mm were observed at a range of power densities and therefore, required the addition of filler metal to avoid the crack sensitive composition.

Positive defocusing was beneficial when welding with filler wire, as it was incident with the laser beam impingement point. The weld surface appearance improved significantly with a smoother surface. On the other hand, large positive defocusing without filler wire caused cracking due to excessive weld width where the gap formed during welding was insufficiently backfilled to heal the initiated crack, so negative or no defocusing produced better weld quality in autogenous welds by increasing the weld pool backfill during solidification.

The extent of Mg vaporisation was a function of welding speed, which decreased with increasing welding speed due to short. Porosity was not observed at higher welding speeds as expected, either because the keyhole stability increased with welding speed or there was not enough time for nucleation.

Increasing the total heat input or laser power to more material being melted so the contraction strain became larger and caused larger cracks to form. Also, a slower solidification rate due to higher laser power caused impurity elements to diffuse into the molten pool and the resulting weakened microstructure of the weld in the middle was susceptible to cracking. A sufficiently low welding speed together with a relatively low heat input or laser power was required to minimise the risk of transverse cracking that is caused by an elongated temperature distribution in the welding direction. Wider welds were found to be less crack sensitive than narrow welds so a lower welding speed, on average produced higher quality welds provided a sufficiently high-power density.

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