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Laser Welding of Magnesium Alloys[†]

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Abstract

Aiming at establishing a reliable welding technique for magnesium alloys, the laser weldability of different types of magnesium alloys has been investigated using a high-brightness disc laser. It was found that high quality joints could be fabricated in the case of wrought type alloys when a certain defocusing level was optimized for the employed high-brightness laser beam of the newly developed diode-pumped solid-state laser. Porosity formation in the case of die-cast type alloys was extensively studied, and it was revealed that the gases entrapped in the base material during the die-casting process grew much larger. It was also found in the case of powder metallurgy alloys that the base material produced an extremely porous weld metal due to the gases entrapped in the production processes. While a significant reduction in porosity formation could be realized by the insertion of extruded alloy sheet in butt welding of die-cast alloys, the higher pressure levels applied in the production of the powder metallurgy type alloys made fusion welding of this alloy very difficult.

KEY WORDS: (magnesium alloys), (laser welding), (disc laser), (die-cast magnesium alloy), (powder metallurgy magnesium alloy)

1. Introduction

The strict environmental regulations as well as the high competence to manufacture more convenient and economical products have caused the attention to be drawn towards magnesium alloys in different fields of industry including aerospace, automotive and electronics¹, ²⁾. This is attributed to the prominent material properties of these alloys such as light weight, superior specific strength, good electromagnetic shielding, good castability, high heat dissipation capability and excellent recyclability³⁻⁵⁾. A wider application of magnesium alloys the aforementioned industries requires in the development of reliable welding methodologies capable of producing reproducible high quality joints between similar magnesium alloys and between magnesium alloys and other structural materials.

Conventionally, magnesium alloys are welded by gas metal arc welding (GMAW) and gas tungsten arc welding (GTAW) processes. However, the characteristic low welding speeds and high heat input of the arc welding processes result in defects including wide heat affected zone (HAZ), high residual stresses and distortion, which makes laser welding a prefered choice⁶⁾. The majority of the reported experiments on laser welding of magnesium alloys were conducted using CO₂ or Nd:YAG laser sources. Magnesium alloys were reported to have higher absorptivity to the near-infrared radiation of the Nd:YAG laser compared with the far-infrared CO₂ laser beams. This resulted in lower threshold irradiance required for the formation of a keyhole and better process stability⁷⁾. In addition, the pronounced negative influence of the plasma, formed during keyhole welding, in the case of the CO₂ laser necessitates helium that has high ionization potential and good thermal conductivity to be used as a shielding gas. While higher consistency in the welded joints strengths can be obtained with Nd:YAG laser regardless of the employed shielding gas⁸⁾. This indicates that magnesium alloys have better weldability with the near-infrared laser sources. Nevertheless, the low wall plug efficiency (1 - 4 %) and beam quality are major disadvantages of lamp-pumped Nd:YAG lasers⁹⁾.

In this article we report our experimental results on welding magnesium alloys with a disc laser that combines the advantages of high beam quality and near infrared radiation^{10, 11}. The experiments were extended to investigate the welding characteristics of different alloy types including wrought, die-cast and powder metallurgy alloys.

2. Experimental

The base materials used in this study were AZ31B and AZ91D extruded Mg alloys, AZ91D die-cast Mg alloy and AZ31B powder metallurgy (P/M) Mg alloy. The chemical compositions are given in **Table 1**. Welding was performed with a 16 kW continuous-wave disc laser beam of 1.03 µm wavelength and 8 mm*mrad

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materials.				
Туре	Al	Zn	Mn	Mg
Extruded AZ31B	3.28	0.81	0.29	Bal.
Die-cast AZ91D	8.8	0.64	0.155	Bal.
Extruded AZ91D	8.9	0.78	0.23	Bal.
P/M AZ31B	3.14	1.08	0.39	Bal.

 Table 1 Chemical compositions (mass %) of base materials.

beam parameter product (BPP). During welding, the top surfaces of the specimens were shielded by Ar gas flowing at 30 L/min through a 16 mm diameter nozzle directed in the welding direction at an angle of 50° with the specimen. The bottom surfaces were also shielded by Ar gas at a flow rate of 20 L/min through the holding fixture. To analyze and understand the different physical phenomena in laser welding of Mg alloys, the molten pool was monitored during welding with a high speed video camera and a high speed X-ray transmission real-time imaging system.

Types and compositions of gases entrapped inside the pores or porosity were identified using a quadrupole mass analyzer. The technique involves drilling holes of approximately 2 mm diameter inside the material containing pores under high vacuum (4×10^{-6} Pa). The mass spectra of the released gases are then analyzed by a Q-mass spectrometer. Samples for metallographic examinations were cut across the welded joints, mounted and ground with SiC paper. Final polishing was performed using 0.25 µm diamond spray. Specimens of extruded AZ31B and die-cast AZ91D alloys were etched with a solution composed of 5 mL acetic acid, 10 mL water, 6 g picric acid and 100 mL ethanol. P/M AZ31B alloy was etched with a solution composed of 25 mL acetic acid, 25 mL water, 10 g picric acid and 175 mL ethanol. Macro- and micro-structures of the welded joints were examined with an optical microscope (OM). Further investigations on the microstructures were carried out using a scanning electron microscope (SEM) coupled with an energy dispersive X-ray spectroscopy (EDS) for chemical compositions analysis. the Vickers microhardness measurements were conducted across the joint cross section at load of 100 gf for 15 s dwell time. The standard tensile test was performed by applying a load transversely to the weld bead at room temperature and a constant cross head speed of 1 mm/min according to ASTM standard¹²). Three specimens were tested for each investigated case.

3. Results and Discussion

3.1 Welding of wrought Mg alloy

Bead-on-plate welding was carried out on 2.8 mm thick plates of AZ31B extruded Mg alloy. The influence of power density was investigated by varying the defocusing distance from -6 mm to +6 mm with an increment of 3 mm and by keeping the laser power constant at 5 kW at the welding speed of 5, 10, 15 and 20

m/min.

3.1.1 Influence of different processing parameters

Typical examples of macroscopic appearances of bead surfaces and fusion zones of AZ31B alloy welded under different defocused distances are shown in **Fig. 1**. Generally, all specimens exhibited high aspect ratios of fusion zones which are characteristic of a keyhole mode welding.

With a tightly focused laser beam at the defocused distance of 0 mm, a small-sized molten pool was formed at the level of about 1 mm length measured in the welding direction. The welding process was unstable and an excessive number of spatters were produced. This resulted in irregular bead shapes and underfilled welded joints. Humping was formed at a welding speed of 15 m/min or higher caused by higher resultant surface tension force due to a narrow weld bead of about 0.5 mm width. By defocusing the laser beam, the stability of the welding process was relatively improved, and the best result was obtained at the defocused distance of -3 mm. Partially penetrated welds were obtained at the defocused distances of ± 6 mm and the welding speeds of 10 m/min or higher. Whilst smooth bead appearances with regular ripples were obtained, undercuts were present.

The influences of laser power and welding speed were evaluated in the second set of experiments where the defocused distance was set at -3 mm, the laser power was varied from 2 kW to 6 kW with an increment of 1 kW, and the welding speed was varied from 2 m/min to 20 m/min with an increment of 2 m/min. Typical examples of macroscopic appearances of bead surfaces and fusion zones are given in **Fig. 2**. As the energy input increased (lower welding speed and higher laser power), larger fusion zones were obtained. However, no obvious relationship between process parameters of laser power or welding speed and soundness of the welded joint could be derived.



Fig. 1 Surface appearances and macrostructures of welded joints at 5 kW laser power and different defocused distances and welding speeds.



Fig. 2 Surface appearances and cross-sectional macrostructures of weld beads produced at -3 mm defocused distance and different laser powers and welding speeds.

3.1.2 Analysis of high speed video camera observations

During all welding experiments, the molten pool was monitored by high speed video camera at 10000 F/s framing rate to evaluate the influence of different process parameters on keyhole stability. Observation examples of a molten pool during formation of a sound weld are given in **Fig. 3**. The specimen was welded at 5 kW laser power, 5 m/min welding speed and -3 mm defocused distance. Melted metal was flowing smoothly behind the keyhole into the molten pool. The keyhole and the molten pool were stable.

High speed video observation results of a specimen welded at 5 kW laser power and 5 m/min welding speed without defocusing of the laser beam are shown in **Fig. 4**. The keyhole inlet was unstable and its size was highly fluctuating. This resulted in turbulent flow of the melted metal and disturbed the molten pool. Moreover, the ejection of the metal vapor out from the keyhole was interrupted by keyhole inlet fluctuation which caused the



Fig. 3 High speed images of stable molten pool and keyhole.



Fig. 4 Formation of spattering and underfilling.

metal vapor to accumulate inside the keyhole. When its vapor pressure exceeded a certain limit, the molten metal near the keyhole inlet burst, leading to spattering of some melted metal droplets.

As welding was autogenously performed without filler wire, the expulsion of the molten metal in the form of spattering resulted in an underfilled weld bead. Just after the laser beam moved out from the underfilling region, a new molten pool began to form. Due to gravity force, some of the newly melted metal flowed back into the underfilling. The amount of this escaping molten metal was too small to refill the underfilled region. However, with the continuous movement of the laser beam, there was not enough melted metal around the keyhole to form a molten pool with the same size as the one before the underfilling. This resulted eventually in non-uniform bead width.

3.1.3 Weld metal characteristics

As can be seen from Fig. 1, the frequently produced welding defects are underfilling, undercuts, root concavity and spatters. By applying the stringent criteria of ISO 13919-2 standard¹³, we could construct a process diagram including the whole range of the investigated parameters. This diagram was first divided into two regions representing full and partial penetration welds. Then, the full penetration region was further divided into two areas of sound and defected welds, as can be seen in Fig. 5. Two points L and H representing different values of energy input were selected in the sound welds region of the process diagram. Point L (lower energy input) corresponds to 4.5 kW laser power and 6 m/min welding speed. Point H (higher energy input) corresponds to 5.5 kW laser power and 3 m/min welding speed. Specimens were welded in butt joint configuration according to the conditions L and H, and hereafter referred to as joint L and joint H, respectively.



Fig. 5 Process diagram of laser welding of AZ31B extruded Mg alloy.

(1) Mechanical properties

The hardness profiles across the welded joints are shown in **Fig. 6**. Joint L exhibited an average hardness of 56.3 ± 0.7 HV 100 gf which was almost the same as that of the base material: 57.3 ± 0.5 HV 100 gf. Increasing the energy input caused the hardness to slightly decrease to



Fig. 6 Hardness distribution across the fusion zone of AZ31B alloy welded joints.

 53.3 ± 1.1 HV 100 gf in joint H.

Regarding the tensile testing, fracture always occurred in the weld fusion zone. The tensile test data given in **Table 2** reveal that joint L possessed a similar tensile strength to that of the base material, while joint H had a tensile strength up to 90.5% of that of the base material. There was a sharp drop in elongation.

(2) Microstructures

The microstructure of AZ31B base material shown in Fig. 7 (a) consists of α -Mg matrix with particles of Mn-Al compound and fragmented β -Mg₁₇Al₁₂ precipitates³⁾. The α -Mg grains are equiaxed with nonuniform size which is typical for extruded structures. Due to the high power density of the laser beam and the high thermal conductivity of Mg-alloys, there was no clear grain coarsening in the HAZ, as exhibited in Fig. 7 (b) and (d). Inside the fusion zone, the microstructure comprised a columnar grains region next to the fusion line, as also shown in Fig. 7 (b) and (d), and a fine equiaxed grains region in the center of the fusion zone, as given in Fig. 7 (c) and (e). The columnar grains region

 Table 2 Tensile properties of base materials and welded joints.

	5			
Tensile specimen	Yield strength, MPa	Tensile strength, MPa	Joint efficiency, %	Elongation, %
Base material	145.5	272.5		20
Joint L	140.7	265.7	97.5	10.7
Joint H	134	246.7	90.5	8



Fig. 7 Optical micrographs of AZ31B alloy: (a) base material, (b) joint L fusion line, (c) joint L weld metal, (d) joint H fusion line, (e) joint H weld metal and (f) EDS spectrum of precipitates.

was formed by epitaxial growth due to the large thermal gradient at the fusion boundary. On the other hand, in the center of the weld fusion zone (FZ), the nucleation was homogeneous and there were no such high thermal gradients. Therefore the grains were equiaxed. Compared with the microstructure of the base material, the weld metal contained coarser and higher volume fraction of precipitates. It has been reported that some of Al atoms in these precipitates can be replaced by Zn and the intermetallic compounds have a form of $Mg_{17}(Al,Zn)_{12}^{12}$ This was confirmed by EDS analysis, as shown in Fig. 7 (f). In addition, the average grain size was smaller in the FZ than that in the base material due to the high-speed melting and rapid solidification of the laser welding process, as exhibited in Fig. 7 (a) and (c). The influence of increased energy input on the microstructure of the FZ is readily apparent in Fig. 7 (e), where larger grains are present.

The mechanical properties, mentioned in section 3.1.3.1, can be analyzed in relation to the microstructure as follows. On one hand, the cast microstructure of the weld metal tends to degrade the mechanical properties. On the other hand the grain refinement tends to improve them. The same levels of hardness and tensile strength of joint L compared with the base material might be attributed to the compensation for the loss of work hardening in the cast microstructure of the weld metal by grain refinement. This was confirmed when coarse microstructure was present in joint H by increasing the energy input which led to lower hardness and tensile strength. The sharp drop in the elongation might be attributed to the cast microstructure of the FZ with hard and brittle micro segregated intermetallic compounds¹⁵.

3.2 Welding of die-cast Mg alloy

It was reported to be difficult to obtain sound joints in laser welding of die-cast Mg alloy because porosity was inevitably produced in the weld fusion zone⁶⁾. In this part we will introduce the methodology we adopted to analyze the mechanism of porosity formation and to find a suitable remedy. Bead-on-plate welding was conducted on 3 mm thick plates of AZ91D die-cast Mg alloy. An optical micrograph of as-polished cross section of the base material is shown in **Fig. 8**. It is observed that the



Fig. 8 Typical macroscopic appearance of as-polished cross section of die-cast AZ91D magnesium alloy.

material contains a significant amount of small-sized porosity, especially in the middle of the plate. The average porosity content in the base material is found to be approximately 2%.

3.2.1 Bead-on-plate welding results

Figure 9 shows some typical examples of fusion zone macrostructure of weld beads produced at different laser powers and welding speeds. A large amount of porosity was formed in all welded specimens, particularly in the weld fusion zones produced at high laser powers or low welding speeds. M. Marya and G.R. Edwards indicated that shielding configuration had a substantial influence on porosity formation during laser welding of high-pressure die-cast magnesium alloys¹⁶⁾. The inadequacy of the shielding allowed the surrounding air to interact with the molten magnesium resulting in significant porosity. However, the employed shielding configuration has already been used in laser welding of wrought magnesium alloys and porosity was hardly formed. additional welding experiments Moreover, were conducted using a gas chamber and the same results were obtained. Porosity in the FZ was mainly in the spherical geometry. Some porosity near the fusion boundary was elongated or protuberant indicating the possibility of being formed due to expansion of gases present originally in the pores in the base material.

The influence of laser power and welding speed on porosity formation is represented in **Fig. 10** where porosity areas in the fusion zones are plotted as a function of welding speed at different laser powers. It is obvious that the weld metal porosity decreased with increasing the welding speed or decreasing the laser power. Basically, increasing the welding speed or decreasing the laser power would lead to the decrease in the fusion zone size as illustrated in **Fig. 11**. These results indicate that weld



Fig. 9 Cross-sectional macrostructure of weld beads in AZ91D die-cast magnesium alloy produced at laser power and welding speed of (a) 4 kW and 4 m/min, (b) 3 kW and 4 m/min, (c) 2 kW and 6 m/min and (d) 1 kW and 8 m/min, respectively.



Fig. 10 Porosity area as a function of welding speed.



Fig. 11 Fusion zone size as a function of welding speed.

metal porosity and fusion zone size are simply related. This behavior supports the conclusion derived from the macrostructural observations as well as the previous results of 17, 18, 19 and 20) suggesting that the weld metal porosity was evolved from the pre-existing pores in the base material, because a larger fusion zone size means that more base material porosity is included and more gases are released during welding.

3.2.2 X-ray transmission real-time observation of keyhole and molten pool

In the current experiments, the keyhole behavior and

porosity evolution inside the weld pool were observed by high-speed X-ray real-time imaging system. It was observed that once a laser beam was irradiated and a keyhole was formed, many bubbles were generated in the weld pool just behind and along the rear wall of the keyhole. While bubbles are being generated, the rear wall of the keyhole was clearly identified. This means that bubbles were not generated from the keyhole. However, a bulge was intermittently formed on the rear wall of the keyhole. This bulge was moving fast on the keyhole wall and occasionally detached in the form of bubbles. Besides, some bubbles were generated from the keyhole tip in the bottom of the weld pool due to keyhole instability. Once formed, the bubbles drifted back into the weld pool by the vigorous melt flows that have been induced by the recoil pressure of evaporation⁹⁾. Then, the fast advancing solidification front, characteristic of laser welding¹⁶⁾, captured the bubbles before floating upward to the weld pool surface under the effect of buoyancy, which eventually resulted in porosity. Snapshots from the X-ray real-time observations during laser welding are shown in Fig. 12, illustrating the situation inside the weld pool. In the first snapshot at time t_1 , many bubbles could be identified just behind the keyhole and along the depth of the molten pool. The second two snapshots indicate the generation of a bubble from a bulged keyhole. At time t_1 + 0.172 s, a bulge was formed on the rear wall of the keyhole. After 8 ms that bulge was detached from the keyhole to form a bubble, and then drifted back into the molten pool.

Another example of the X-ray real-time observations is given in **Fig. 13**. Two bubbles were identified and marked with the numbers 1 and 2 at time t_1 (0.032 s before switching the laser beam off). As time progressed, bubbles 1 and 2 grew in size and channels (referred to by yellow arrows) connecting them to the fusion boundary could be seen as shown in the snapshot at time $t_1 + 0.032$ s. Also, it was observed that bubbles were growing in a direction parallel to these channels towards the inside of the molten pool (away from the fusion boundary). This indicates that bubble growth was caused by swelling of the expanding gas pre-existing in the base material. Bubbles continued to grow even after the laser beam was



Fig. 12 Porosity evolution during laser welding of die-cast AZ91D magnesium alloy.



Fig. 13 Bubbles growth and coalescence.

switched off and eventually coalesced into one large bubble as shown at $t_1 + 0.06$ s.

3.2.3 Analysis of gas compositions inside porosity

Gases retained inside the base and weld metal porosity were analyzed at two different locations (referred to as point 1 and point 2) for each specimen. The analysis results are given in Table 3. It is evident that the main gas constituent in the base material and weld metal porosity is nitrogen. This would imply that nitrogen gas (air) was entrapped during die-cast processing and led to the formation of the base material porosity. The nitrogen gas swelled in the high-temperature molten pool during welding, resulting in weld metal porosity. It is also notable that the percentage of hydrogen in the weld metal porosity is higher than that in the base material porosity. This is interpreted by considering that hydrogen is rejected at the solid-liquid interface because the solubility of hydrogen in the solid magnesium phase is lower than that in the liquid magnesium phase¹⁶, and that the rejected hydrogen is included in the pre-existent bubble as soon as the advancing solidification front encounters the bubble. It is also believed that metal vapors entrapped in the molten pool bubbles due to keyhole instability are condensed on

Table 3 Gas compositions (%) in porosity.

Cas component	Base material		Weld metal	
Gas component	Point 1	Point 2	Point 1	Point 2
$H_2 (M/Z^a 2)$	0.9	0.8	16.6	4.2
CHG ^b (M/Z 15)	2.1	1.9	5.0	1.1
N ₂ (M/Z 28)	95.3	95.0	78.2	94.1
Ar (M/Z 40)	0.6	0.7	0.2	0.6
CHG (M/Z 43)	1.0	1.2		
CO ₂ (M/Z 44)	0.1	0.3		
CHG (M/Z 57)		0.1		

^a M/Z refers to mass-to-charge ratio

^b CHG refers to hydrocarbon gas

the walls of pores as the temperature decreased after welding. This lowers the pressure inside the porosity and allows hydrogen to diffuse from the weld metal to porosity. Thus the hydrogen content of the porosity in weld fusion zone should increase during and after laser welding. These results are different from those reported in^{21, 22)} where porosity resulting from keyhole instability was found to mainly include helium (shielding gas) and hydrogen gas. This means that keyhole instability is not the major concern.

Based on these results, it can be concluded that gases entrapped inside the pre-existing porosity of the base material were released during welding. This, in addition to keyhole instability, resulted in bubbles formation in the molten pool. The presence of vigorous melt flows and the rapid solidification associated with the laser welding process prevented these bubbles from escaping out from the weld pool, resulting in porosity formation.

3.2.4 Porosity reduction approach

Most of the published reports on laser welding of cast Mg allovs have been focused on studying the characteristics of the welding process and very few attempts have been devoted to propose and analyze a mechanism to prevent or reduce porosity formation. M. Marya and G.R. Edwards¹⁶⁾ studied the influence of interaction time (beam diameter/welding speed) on pore formation and concluded that short interaction time (small beam diameter and high welding speed) can minimize pore formation because there will be insufficient time for pores to nucleate and grow. However, reduction of porosity by employing high welding speed would be at the expense of penetration as can be seen from Fig. 9 (d). H. Zhao and T. DebRoy²⁰⁾ proposed remelting of the fusion zone by a second welding run and explained that the large pores in the fusion zone of the first run might have a second chance to float out of the weld pool of the second run. Nevertheless, the parameters of the second run should be carefully controlled because the first welding run changed the surface properties and this would, in turn, influence the laser-material interaction. Thereby, if the second run was performed with the same parameters as that of the first run, it would result in the formation of larger fusion zone and new porosity to be introduced in the weld bead.

Laser Welding of Magnesium Alloys

As discussed in the previous sections, the weld metal porosity associated with laser welding of die-cast magnesium alloys originated mainly from the pre-existing pores in the base material and partly from the keyhole instability. Full penetration welding has been proved to be effective in preventing porosity related to keyhole instability as there would not be bubbles coming from the keyhole tip²¹⁾. On the other hand, minimizing the portion of the base material to be melted during welding might be a good measure to reduce the weld metal porosity coming from the pre-existing pores in the base material. Accordingly, butt joints of AZ91D die-cast magnesium alloy were designed with an insert sheet of extruded porosity-free AZ91D magnesium alloy comprising the same chemical compositions.

Figure 14 displays some typical examples of fusion zone macrostructure of butt-welded joints with the insert sheets produced under the optimized conditions. The solid and dashed yellow lines represent the original boundaries of the die-cast alloy and the fusion zone boundaries, respectively. The welding parameters were optimized to minimize the volume of melted base material that was estimated by measuring the confined area between the solid and dashed yellow lines.



Fig. 14 Fusion zone macrostructure of butt joints produced in die-cast AZ91D alloy with insert-layer at 3 kW laser power and 5 m/min welding speed.

To have a consistent comparison between welding with insert-layer sheet and autogenous welding, the area percent porosity was measured in joints welded with optimized welding parameters in the case of welding with insert-layer and joints welded with minimum energy input required for full penetration in the case of autogenous welding. **Figure 15** depicts such a comparison. The use of an insert-layer sheet could reduce the porosity in the weld metal to levels close to that of the pre-existing porosity in the base material and 50 % less than that of the autogenously welded joints²³.



Fig. 15 Area percent porosity in base die-cast AZ91D alloy, and joints welded with or without insert-layer.

3.3 Welding of powder metallurgy Mg alloy

The poor room temperature formability of magnesium alloys is still limiting their widespread applications²⁴. Therefore, various techniques including shear rolling²⁵, addition²⁶⁾, earths equal-channel rare angular processing²⁷⁾ and rapid solidification²⁸⁾ have been investigated to overcome this problem. The objective of the present experiments is to investigate the laser weldability of rapidly solidified powder metallurgy (P/M) magnesium alloys. The initial powders were prepared by spinning water atomization process (SWAP)²⁹; a process capable of producing powders with very fine grained microstructure as can be seen in Fig. 16. 3 mm thick specimens of hot extruded P/M AZ31B magnesium alloy were subjected to bead-on-plate welding under different conditions.

3.3.1 Bead-on-plate welding results

Figure 17 shows some typical examples of fusion zone macrostructure of weld beads produced at different laser powers and welding speeds. An extremely porous fusion zone was formed in all welded specimens. Variation of laser welding parameters influenced only the fusion zone size and did not affect the porosity formation propensity. A large number of micro and macro pores with spherical or irregular shapes were formed under all



Fig. 16 Microstructure of P/M AZ31B alloy base material.



Fig. 17 Cross-sectional macrostructure of weld beads produced at laser power and welding speed of (a) 4 kW and 4 m/min, (b) 3 kW and 6 m/min, (c) 2 kW and 8 m/min and (d) 1 kW and 10 m/min, respectively.

the welding conditions. This abnormal amount of porosity displaced the weld metal resulting in large excess weld metal and excessive penetration, as indicated in Fig. 17 (b) and (c). Moreover, the welded joints were so weak that they could be easily broken by hand. It should be noted that the shielding configuration did not have any influence as the same results were obtained in the case of shielding nozzle as well as in the case of gas chamber.

Investigation of the fusion zone microstructure revealed that many micro pores were present near and along the fusion boundary, as shown in **Fig. 18**. Also, it was observed that some pores near the fusion boundary



Fig. 18 Pores and their morphology near fusion boundary.

were elongated and featuring a growth direction towards the inside of the fusion zone starting from the fusion boundary. Two examples of such pores are marked with white arrows in Fig. 18. This pore morphology suggests similar behavior to that of the die-cast alloy, i.e. the porosity was formed by infusion of gases from the base material. However, the base material microstructure did not contain any porosity (Fig. 16). Also, the observed pores and weld metal displacement are much larger than those in the case of the die-cast alloy. Therefore, more in-depth investigations were carried out to reveal the mechanism of this abnormal pore formation and consequently to find a suitable measure.

3.3.2 X-ray transmission real-time observation of keyhole and molten pool

During laser beam welding, the keyhole behavior and porosity evolution inside the weld pool were observed using a high-speed X-ray real-time imaging system. It was observed that once a laser beam was irradiated and a keyhole was formed, many bubbles started to emerge from the weld pool. Bubbles could be identified only at the upper part of the weld pool. No melt flows were observed to drift the bubbles backwards in the weld pool instead; bubbles displaced the weld metal on their way out from the weld pool. The molten metal was displaced to a height of approximately 4 mm over the specimen surface while the keyhole penetrated to a depth of approximately 6 mm. Some bubbles could escape out from the weld metal but many others were captured by the rapid solidification resulting in porosity. Snapshots from the X-ray transmission real-time observation results during laser welding, taken at 0.048 s, 0.104 s and 0.664 s after the laser beam was switched on, are shown in Fig. 19, illustrating the process of bubbles generation and



Fig. 19 Bubbles generation and molten metal displacement.

weld metal displacement. At time t = 0.048 s the height of the displaced molten metal was approximately 2 mm. This height was increased to 4 mm at time t = 0.104 s. As time progressed, the situation remained unchanged; generation of many bubbles and displacement of the molten metal, as shown at time t= 0.664 s. Also it was observed that neither the rear wall of the keyhole nor bubbles coming from the bottom part of the weld pool could be identified. This is believed to be due to the presence of large volume of gases and a small amount of molten metal in this part of the weld pool.

3.3.3 Analysis of gas compositions inside porosity

Gases retained inside the weld metal porosity were analyzed at two different locations referred to as point 1 and point 2. The data are given in **Table 4**. The analytical results reveal that the porosity contained only nitrogen gas and a very small amount of argon. Recalling that welding inside a gas chamber did not prevent the porosity formation, this nitrogen is believed to come from the base material. It is also notable that hydrogen was not detected in the porosity. Hydrogen rejection at the solid-liquid interface can contribute to porosity formation at the last solidification stages of magnesium alloys and some hydrogen had already been detected inside the porosity of the die-cast AZ91D alloy welds. However, hydrogen rejection was reported to occur at the Mg₁₇Al₁₂ intermetallic compound in the alloy AZ91 (containing approx. 9 wt.% Al)³⁰⁾. Besides the difference in hydrogen solubility between the liquid and solid phases in magnesium is smaller than that in aluminum³¹ which means that hydrogen rejection is less in the case of magnesium. Therefore, the absence of Mg₁₇Al₁₂ phase in the rapidly solidified P/M alloy microstructure (as will be indicated shortly) might be the reason why hydrogen was not detected in the porosity.

Table 4 Gas compositions (%) in porosity.

Gas component	Point 1	Point 2
$H_2 (M/Z^a 2)$		
CHG ^b (M/Z 15)		
N ₂ (M/Z 28)	100	99.7
Ar (M/Z 40)	< 0.1	0.3
CHG (M/Z 43)		
CO ₂ (M/Z 44)		
CHG (M/Z 57)		

^a M/Z refers to mass-to-charge ratio

^b CHG refers to hydrocarbon gas

3.3.4 Chemical compositions analysis of base material

Figure 20 shows a SEM image with some EDS spectral analyses of the base material. In general, the microstructure of a conventionally extruded AZ31B



Fig. 20 Microstructure and EDS analysis of the base material, (a) SEM photo of the microstructure indicating the locations for the analysis, (b) EDS spectrum for the matrix (spectrum 1 in (a)) and (c) EDS spectrum for the white particles (spectrum 2 in (a)).

magnesium alloy consists of α -magnesium matrix and intermetallic compound β -Mg₁₇Al₁₂³⁾. However, the very rapid solidification rates associated with the SWAP retard the precipitation of the intermetallic compounds and produces super saturated powders, as reported in 28). Figure 20 (a) indicates that the microstructure comprises a fine grained matrix and white particles aligned in the extrusion direction. As can be seen from Fig. 20 (b) and (c), nitrogen was not detected in neither the matrix nor the white particles and these particles contained mainly magnesium and oxygen. Accordingly, the base material microstructure could be considered to consist of a super saturated α -magnesium matrix containing particles of magnesium oxide and no nitrogen-containing compounds could be detected.

To clarify the source of the nitrogen gas detected in the weld metal porosity, a specimen was prepared from the base material and tested in a gas analyzer where it was melted and the evolved gases were analyzed. The results are given in **Table 5**. A specimen of conventionally extruded AZ31B alloy was also examined

 Table 5 Nitrogen and oxygen contents (ppm) in the base material.

Cracimor	Oxygen		Nitrogen	
Specimen	Mean	STD	Mean	STD
Extruded-P/M AZ31B	14.95	1.8	24.93	0.5
Extruded-cast AZ31B	6.2	0.3	16.4	1.3

for comparison. It is seen that the extruded-P/M alloy contained slightly higher amounts of oxygen and nitrogen compared with the extruded-cast alloy. It is believed that these gases were contained in the base material in the initial production processes since the green compacted powder had a relative density of 92%. Besides, SWAP powders were reported to contain some oxides²⁹⁾. Due to its higher affinity to magnesium, oxygen reacted with magnesium forming oxides while nitrogen remained in an uncombined state inside the base material since it was not detected by EDS analysis. This would also explain the reason why only nitrogen gas was detected in the weld metal porosity. Once the base material is melted during welding, nitrogen and part of the oxygen are released forming bubbles in the molten metal. Oxygen reacts with magnesium whereas nitrogen is retained inside the weld metal porosity.

Figure 21 shows SEM photos of the fractured surfaces of a welded joint and the corresponding EDS spectra. Small white particles were observed inside and outside the pores, as seen in Fig. 21 (b) and (c). The EDS spectrum in Fig. 21 (d) indicates that these white particles are magnesium oxide. These particles, most probably, were formed by the reaction of oxygen with magnesium from the metal vapor. Oxygen was also detected on the fractured surface inside and outside the pores, as shown in Fig. 21 (e).

Based on these results it could be concluded that when the material was melted during welding, nitrogen that was initially entrapped in the base material during the production processes was released and led to the formation of large pores in the weld metal. However,

such a small amount of nitrogen alone detected in the base material cannot explain the abnormally large pores observed in the weld metal since particularly porosity was hardly formed in the extruded-cast alloy welds, although the difference in gas contents between the two materials is not so high. The applied pressure during powder compaction and extrusion processes has to be taken into consideration to adequately analyze the abnormal pore formation upon welding. The initial SWAP powder was compacted under a pressure of more than 3800 atm. and then hot extruded with an extrusion ratio of 42. This means that the pressure of any gases entrapped inside the material would be very much higher than the atmospheric pressure. The liberated gas during welding expands to release the pressure forming large pores and displacing the molten metal out from the weld pool. Another factor that might have magnesium as I. J. Polmear indicated that hydrogen is the only gas that dissolves in molten magnesium³¹⁾.

3.3.5 Porosity reduction approach

The above mentioned results indicate that the weld metal porosity originated from the pre-existing gases in the base material. Therefore, minimizing the portion of the base material to be melted during welding might be a good measure to reduce the pore formation. Accordingly, butt joints of extruded-P/M AZ31B magnesium alloy were designed with an insert-layer of extruded-cast AZ31B magnesium alloy comprising the same chemical compositions. Despite there is no big difference in the gas contents between the two materials, this approach might help to reduce the porosity because gases inside the



Fig. 21 Fractured surfaces and EDS analysis of a welded joint, (a) SEM photo of the fractured surface, (b) magnified view of the black boxed region in (a), (c) magnified view of the white boxed region in (a), (d) EDS spectrum for the white particles (spectrum 1 in (b) and (c)) and (e) EDS spectrum for the fracture surface (spectrum 2 in (b) and (c)).



Fig. 22 Macrostructural weld fusion zones of butt joints with insert-layer in extruded-P/M AZ31B magnesium alloy produced at 3 kW laser power and (a) 4 m/min, (b) 6 m/min and (c) 8 m/min welding speed.

extruded-cast alloy are not retained under high pressure. Besides, the approach has already been applied in the case of die-cast alloy and the porosity was significantly reduced.

Figure 22 displays some typical examples of fusion zone macrostructure of butt-welded joints with the insert-layer. In general, a smaller number of pores were formed compared with autogenous welds. However, the pore proportion is still high as well as the excess weld metal and the excessive penetration, as seen in Fig. 22 (a). A reduction in the energy input by increasing the welding speed led to a decrease in the melted part of P/M alloy and consequently a reduction in the weld metal porosity, as observed in Fig. 22 (b) and (c). Nevertheless, the energy input was not sufficient to achieve adequate fusion. Moreover, even in this minimized melted part of the P/M alloy significant porosity was formed particularly near and along the fusion boundary. The difference in the effectiveness of the porosity reduction approach between this case and the die-cast alloy case might be attributed to the higher pressure levels of the entrapped gases of the P/M alloy. These results confirm the proposed mechanism of porosity evolution on one hand, but on the other hand, indicate that it is very difficult to produce sound welded joints in the material under consideration using fusion welding 32 .

4. Summary

A high power and high brightness disc laser was used in similar welding of extruded AZ31B, die-cast AZ91D and P/M AZ31B Mg alloys. In the case of the wrought AZ31B alloy, observation of the molten pool with a high speed video camera during welding elucidated the prominent influence of the power density of the laser beam on the stability of the keyhole and the resultant weld quality. For example, under the conditions of a tightly focused laser beam and low welding speed, spattering was caused by the variation in metal vapor pressure inside an unstable keyhole due to fluctuations in its inlet size. Stabilization of the keyhole through selection of appropriate power density and other welding parameters could improve the quality of the welded joints. A process window diagram for each alloy that covers a wide range of the processing parameters could be constructed. Sound welds area could be identified on these diagrams where the quality of the welded joints conformed to the relevant standard. Butt welded joints with similar mechanical properties to those of the base materials could be successfully produced.

The problem of porosity formation during laser welding of die-cast magnesium alloys was studied on the AZ91D alloy. A high-speed X-ray transmission real-time imaging system was employed to record the real-time situation inside the molten pool during welding. Many bubbles were observed to be evolved from the base material once melted with a laser beam. The bubbles were mainly formed behind and along the rear wall of the keyhole. Analysis of the gases retained inside the porosity attributed the bubbles formation to the release of gases, initially entrapped inside the base material during the die-casting process, upon welding. The propensity of porosity formation in the welded joints was found to directly increase with the melted part of the base material. A 50% reduction in the porosity content of the fusion zone could be achieved by the insertion of a thin layer of an extruded AZ91D alloy between the parts to be welded of the die-cast AZ91D alloy.

The third type of magnesium alloy investigated in this study was powder metallurgy alloy. A newly developed powder metallurgy AZ31B alloy was used to investigate the weldability of this alloy type. The initial powders were produced by spinning water atomization process that results in very fine microstructure and consequently enhanced mechanical properties. Although the base material featured a fully dense microstructure, extremely porous weld metal was obtained regardless of the employed welding parameters. The high-speed X-ray transmission real-time observations of the molten pool during welding showed intense bubbles evolution soon after laser irradiation. The vigorous upward movement of the bubbles displaced most of the weld metal out of the fusion zone and resulted in very weak joints. Bubbles formation was again attributed to the expansion of gases initially entrapped in the base material. However, the higher pressure levels employed in the production processes of this alloy type caused larger pore sizes and

displacement of the weld metal than the case of the die-cast alloy. Also, this might be the reason why the insertion of a conventionally extruded AZ31B alloy layer between the powder metallurgy AZ31B alloy parts to be welded did not work as it did in the case of the die-cast alloy.

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