CHAPTER 6 CHARACTERIZATION ON HV-EBW OF DISSIMILAR MG-AL-BASED ALLOYS

Chapters 4 and 5 have illustrated that defects causing SC from EBW are related to Al content of Mg-Al-based weldment and its effect will be most conspicuous for DMW. To prevent defects from forming in Mg-Al-based welds, choosing optimal EBW parameters is necessary. Previous research optimized HV-EBW control parameters to be a 30mA beam current, a 120kV accelerating voltage, a 30mm/sec welding speed, and the focus position at bottom, which is used for welding 11mm-thick Mg-Al-based plates in this chapter. Author will not only discussed the difference between SMW and DMW of AZ-series alloys, but also further extend the difference of chemical compositions to explore this technique.

6.1 Tensile Test for SMW and DMW of Mg-Al-Based Alloys

Figure 38 shows that UTS ratio (weldment/matrix) increases with increasing Al content, regardless of which base material, or SC weldment with SMW or DMW is used (except for pure Al and its weldments). The main reason is that the grain of δ phase (solid solution phase) contains more Al atoms by means of solid solution, and then the remnant Al will precipitate the brittle γ phase (Mg_{17}(Al,Zn)_{12}) from the grain boundary [108]. Homogeneity of the base metal is damaged by these precipitates which results in the ductility tending to decrease with increasing Al content. The problem increases when one turns from SMW to DMW. The DMW show lower strength and elongation than the SMW when tested in the as-welded condition.

The strength of NSC weldment increases slightly with increasing Al content if surface stress raiser is removed by milling, which is much higher than that of SC weldment. This implies that the best quality of AZ-series weldments with DMW can be obtained as with SMW. Furthermore, some UTS ratios for NSC weldment are higher than 100% and others range between 90% and 100%. This results from work hardening by milling and more pores in the weld, respectively. On the other hand, the ductility of NSC weldment can improve with original tendency for that of base material. The weld belongs to cast structure, so its
Figure 38  Pictograph of strength and strain of weldment with SMW and DMW for Mg-Al-Based alloys.

elongation is much lower than that of extruded material. However, the average Al content for AZ-series NSC weldments may be ranked in order of decreasing ductility as follows: 4.5wt.%, 6.0wt.%, 7.5wt.%, 3.0wt.%, and 9.0wt.%. Regarding butting pure Mg and Al there is a dramatic decrease for the strength and the ductility of weldment. Further discussions are presented later on.

### 6.2 Microstructure Observation for DMW of Mg-Al-Based Alloys

Figure 39 indicates metallographic photographs of weld cross-sections with SMW and DMW for Mg-Al-based alloys. It can be found that the depth of weld root concavity with DMW decreases with increasing average Al content in the weld (This rule is not applicable to pure Al weldment because the input energy becomes insufficient slightly). This is due to root concavity which is possibly the result of turbulence in the weld pool caused by regions of differing fluidity. Though the melting alloy with higher Al content (or lower liquidus
Figure 39  Comparison of cross sections of welds with SMW and DMW. Plate thickness is 11 mm.

temperature) has better fluidity to fill defects with constant energy in a rapidly cooling weld pool, the turbulence which results from a mixed weld composition can also lead to the formation of numerous pores. The Mg-Al weldment shown in Figure 39(l) is a good example of this. Even if the fluidity of the melting alloy is quite good, the weld cools too quickly to expel all the pores in time.

Figures 40(a)–(c) show metallographic photographs of weld crest boundaries with DMW for the sides of AZ31B, AZ61A, and AZ91D, respectively. Precipitate growth modes are similar to those of SMW. These precipitates concentrate in FZ, in forms from scattered
particles to dense dendrites as Al content increases. Compared with the weld boundary with SMW for the same base material, more particles and less dendrite distribute on either side of weld with DMW. This results from balancing the Al element as shown in Figure 41. The Zn element with low content moves with Al element simultaneously, which contributes to the decreasing of the Gibbs free energy of precipitates [110]. Therefore, part of Al can be replaced with Zn for Mg\textsubscript{17}Al\textsubscript{12}, indicated by Mg\textsubscript{17}(Al,Zn)\textsubscript{12} according to recent research [108]. Additionally, high oxygen content of precipitate also proves the possibility of pore formation as revealed in Figure 40(d).

Figures 40(a)–(c) show that the width of HAZ and the precipitating state of submicron-sized crystal (γ phase) depend on Al content. When Al content is 3.0wt.%, the width of HAZ is about 80–100µm and its average grain size (10µm) is greater than that (5µm) in base metal. After Al content achieves 6.0wt.%, the width is about 30–40µm and its average...
grain size (20µm) is greater than that (15µm) in base material. At this time, submicron-sized crystals are present and distribute over the inside of grain (δ phase) in HAZ. When the Al content reaches 9.0 wt.%, the HAZ can be divided into two parts. One part, HAZ-I, is the partially melted zone (PMZ) beside the FZ [16], just as it was with 6.0 wt.% Al. The other part, HAZ-II, is 50-60µm in width beside the base material without grain growth (15µm), which only precipitates submicron-sized crystals around grain boundary phase. If a 1.5µm-diameter precipitate in HAZ is selected and centered on the precutting area, as shown in Figure 40(d), the specimen can be fabricated by focus Ga ion beam to observe the relationship between γ and δ phases. Figure 42 shows that the precipitate of γ phase is a polycrystalline state (average grain size is about 0.2µm) whose unit cell (BCC structure, \( a \approx 10.6 \text{ Å}, \alpha = \beta = \gamma = 90\degree \)) is much larger than that of δ phase (HCP structure, \( a = b \approx 3.2 \text{ Å}, c \approx 5.2 \text{ Å}, \alpha = \beta = 90\degree, \gamma = 120\degree \)) in terms of the electron diffraction pattern.

When Al and Zn contents reach only 3.0wt.% and 1.0wt.%, respectively, the submicron-sized crystals cannot be precipitated in the grain according to Mg-Al-Zn ternary
Figure 42  TEM photographs and electron diffraction pattern in HAZ for AZ61A: (a) bright field image, (b) electron diffraction pattern with [T 1 1] zone axis for γ phase, and (c) electron diffraction pattern with [0 0 0 1] zone axis for δ phase.

Phase diagram [111]. Therefore, the remnant energy in HAZ can only coarsen grain during EBW. Increasing Al content often causes the γ phase to precipitate continuously along grain boundary under slow cooling. As HAZ temperature is raised between 610°C (liquidus temperature) and 437°C (eutectic temperature) again, many Al atoms diffuse quickly from the liquid-state γ phase to the solid-state δ phase via solid solution. Once the alloy is quickly cooled from eutectic temperature to room temperature, the maximum solid solubility of Al in the precipitate vicinity decreases immediately from 11.5wt.% to about 2.0wt.%. Most Al atoms cannot return to the grain boundary in time, so γ phase will nucleate and grow directly into the submicron-sized crystals by partial HAZ energy at their present positions as shown in Figure 40(d). Therefore, the width of the narrowed grain coarsening zone of AZ61A alloy is slightly wider than the coarsening grain diameter. With regard to AZ91D weldment, its continuous γ phase which extends a longer distance across HAZ than AZ61A weldment can be melted to reform the wider distribution of submicron-sized precipitates during fast solidifying at 437°C. The EBW for AZ61A alloy according to foregoing viewpoints possesses optimal joint efficiency because of the narrowest HAZ.
As the chapters 4 and 5 describes, only five kinds of defects (undercut, root concavity, pores, cracks, and a HAZ) appear in SMW of AZ-series alloys. When butting pure Mg and Al, especially, numerous pores, longitudinal and transverse cracks are found (Figure 43(a)). This greatly weakens the mechanical properties of the weld. Figures 43(b) and 43(c) zoom in on the origin and midcourse (inset) of a crack along the grain boundary. The chemical compositions of positions 44(c) and 44(d) were inspected using an EPMA (Table 41). In these regions the γ and δ phases coexist, as seen in the Mg-Al binary phase diagram. Phases with different thermal expansibilities form simultaneously and are randomly distributed throughout the weld, so the thermal stress distribution is quite complex.

In fact, the composition of this weld can be seen as the epitome of the Mg-Al binary phase diagram. At room temperature there are five possible phases: α-Al (a solid solution), MgAl\(_2\) (a metastable phase), β-Mg\(_2\)Al\(_3\) (an intermetallic compound), γ-Mg\(_{17}\)Al\(_{12}\) (another intermetallic compound), and δ-Mg (another solid solution). The MgAl\(_2\), β, and γ phases are concentrated in the FZ, while α-Al or δ-Mg forms close to base metal. Because pure Al has a higher density and thermal conductivity than pure Mg, the former absorbs more energy during EBW. Molten Al also possesses a larger viscosity than molten Mg, so the former also has much wider PMZ (Figure 44(b)). Figure 44(a) shows the cross-section of a weld between pure Mg and AZ91D. In this case the PMZ is not obvious, because the composition is not very different between both sides of the weld. Also, a low concentration of Zn can replace some of the Al in the γ phase (i.e., Mg\(_{17}\)Al\(_{12}\) becomes Mg\(_{17}\)(Al,Zn)\(_{12}\)) [108], which contributes to the decreasing of the Gibbs free energy of precipitates [110].

In welds butting pure Mg and Al, the distribution of phases is very complex but can be observed by TEM. As shown in Figures 45 and 46, the compositions of neighboring phases are similar; however, there are some needle-like precipitates in the α-Al grain. These phase structures can be identified by analyzing the electron diffraction pattern as the Table 42 shows.

6.3 XRD Analysis for DMW of Mg-Al-Based Alloys

As shown in Figures 47(a) and (e), the (1 0 1 0) plane lying on the transverse plane
Figure 43  Photographs of the microstructure in a Mg-Al weld: (a) Metallographic photograph of the weld cross-section, (b) enlargement of the weld root, (c) SEM image taken near the Al side of the weld root, (d) SEM image taken near the Mg side of the weld root.
of the extruded plate is the preferred orientation before EBW. Weld structure transition from forging to casting after EBW, and the original preferred orientation gradually disappears in curve (c). The HAZ diffraction pattern is a transitional state between base material and FZ, as shown in Figures 47(b) and (d). The γ phase is also the sole intermetallic compound but its peak is not well-marked. Because peak intensity increases with Al content in Mg alloys, it is

Table 41  EPMA data along the Mg-Al weld root [wt.%].

<table>
<thead>
<tr>
<th>Position</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
<th>6</th>
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<tr>
<td>Mg</td>
<td>85.709</td>
<td>77.678</td>
<td>69.395</td>
<td>64.538</td>
<td>60.986</td>
<td>50.240</td>
</tr>
<tr>
<td>Al</td>
<td>14.286</td>
<td>22.311</td>
<td>30.572</td>
<td>35.455</td>
<td>38.994</td>
<td>49.676</td>
</tr>
<tr>
<td>Fe</td>
<td>0.004</td>
<td>0.011</td>
<td>0.033</td>
<td>0.007</td>
<td>0.019</td>
<td>0.084</td>
</tr>
<tr>
<td>Position</td>
<td>7</td>
<td>8</td>
<td>9</td>
<td>10</td>
<td>11</td>
<td>12</td>
</tr>
<tr>
<td>Mg</td>
<td>33.654</td>
<td>61.218</td>
<td>90.754</td>
<td>75.113</td>
<td>83.327</td>
<td>70.071</td>
</tr>
<tr>
<td>Al</td>
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<td>38.722</td>
<td>9.238</td>
<td>24.869</td>
<td>16.673</td>
<td>29.918</td>
</tr>
<tr>
<td>Fe</td>
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<td>0.060</td>
<td>0.008</td>
<td>0.018</td>
<td>0.000</td>
<td>0.011</td>
</tr>
</tbody>
</table>

Figure 44  EPMA analysis of weld crest with DMW for (a)Mg-AZ91D (b)Mg-Al.
Figure 45  TEM inspecting results near Mg side of weld for butting Mg and Al: (a) bright field image, (b) electron diffraction pattern with [0 1 1] zone axis for Mg$_2$Al$_3$, and with (c) [0 1 1] and (d) [1 1 1] for Mg$_{17}$Al$_{12}$.

Figure 46  TEM inspecting results near Al side of weld for butting Mg and Al: (a) bright field image, (b) electron diffraction pattern with [3 3 1] zone axis for MgAl$_2$, and with (c) [1 0 0] and (d) [1 1 1] for α-Al.
Table 42  Composition and structure of phases in the weld butting Mg and Al.

<table>
<thead>
<tr>
<th>Phase</th>
<th>Al content</th>
<th>Structure</th>
<th>Lattice parameter</th>
<th>Space group</th>
<th>Atom per cell</th>
</tr>
</thead>
</table>
| α-Al       | 0-18.6     | FCC       | a = b = c ≈ 4.1 Å  
α = β = γ = 90°  
Fm3m 4 |             |               |
| MgAl₂      | 33.3       | BCT       | a = b ≈ 4.1 Å, c ≈ 26.6 Å  
α = β = γ = 90°  
I41/amd 24 |             |               |
| β-Mg₂Al₃  | 38.5-40.3  | FCC       | a = b = c ≈ 28.2 Å  
α = β = γ = 90°  
Fd3m ~1168 |             |               |
| γ-Mg₁₇Al₁₂| 45-60.5    | BCC       | a = b = c ≈ 10.6 Å  
α = β = γ = 90°  
I43m 58 |             |               |
| δ-Mg       | 69-100     | HCP       | a ≈ 3.2 Å, c ≈ 5.2 Å  
α = β = 90°, γ = 120°  
P6₃/mmc 6 |             |               |

Figure 47  XRD analysis of weld with DMW for AZ31B-AZ91D: (a)base material for AZ31B, (b)HAZ beside AZ31B, (c)FZ, (d)HAZ beside AZ91D, and (e)base material for AZ91D.
only found in Figures 47(d) and (e) after balancing the Al element of weld with DMW.

Curves (a) and (g) in Figure 48 show the XRD patterns of pure Mg and pure Al, respectively. In addition to the original solid solution phases, however, the neighboring HAZ (or PMZ) also contains the MgAl_2, β, and γ phases, as shown in curves (b) and (f). Curves (c) through (e) show XRD patterns in the FZ, where these three phases are most abundant. The peak intensities of these curves indicate that the γ phase is the most abundant, followed by the β phase and MgAl_2. This is because the Gibbs free energy is lowest for the intermetallic phases (γ and β). Not only is the melting point of the γ phase slightly lower than that of β phase, but Mg has higher fluidity and more meltage than Al. The γ phase has the highest Mg content, so will represent most of the weld. As the different phases also have different physical and chemical properties, they cool unevenly and form streaks in the weld.

### 6.4 Microindentation Hardness Test for DMW of Mg-Al-Based Alloys

Figure 49 shows test positions and Vickers hardness of weld cross-section with SMW and DMW for AZ-series alloys. No matter whether by SMW or DMW, Vickers hardness of FZ, HAZ, and base metal increase with increasing Al content. The brittle γ phases for SMW concentrate mainly in the weld crest center causing Vickers hardness dropping from the central line of FZ to base metal and from top to bottom. The DMW precipitates concentrate mainly on the high Al content side of weld crest causing Vickers hardness reduction from right to left and from top to bottom. The vertical and horizontal test results all coincide with analyzed result in Figure 41. Moreover, a sudden decrease of Vickers hardness occurs in HAZ as affected by annealing softening. The situation for DMW is particularly serious, such as (-0.6, 5.0), (-0.5, 0), and (-0.5, -5.0), because these positions become the lowest values along continuously descending hardness. Therefore, the weldment fracture mode for DMW belongs to the regular HAZ fracture, initiating from weld root concavity and propagating along these positions.

Figure 50 shows test positions and Vickers hardness of weld cross-sections with SMW (pure Mg or pure Al) and DMW for Mg-Al-based alloy. The Vickers hardness of pure Mg and pure Al welds are similar, and both have uniform horizontal distributions. In contrast, the
Figure 48  XRD analysis of weld for butting pure Mg and pure Al: (a) pure Mg, (b) HAZ beside Mg, (c) FZ beside Mg, (d) central line of weld, (e) FZ beside Al, (f) HAZ beside Al, (g) pure Al.
curves for DMWs butting pure Mg and AZ-series alloys show a gradient due to the high concentration of brittle precipitates (γ phase) on the high Al content side of weld crest. This causes a gradual reduction of the Vickers hardness from right to left and from top to bottom. A sudden decrease in the Vickers hardness occurs at the HAZ, which is softened by annealing. The situation is particularly serious at positions (-0.4, 5.0), (-0.3, 0), and (-0.3, -5.0), where the hardness is at its minimum. The trend is entirely different in SMWs, where the FZ is actually harder than the base metal. DMWs are transformed from the α and δ phases into a streaky distribution of brittle γ, β, and MgAl₂ phases. Intergranular fracture (propagated by transgranular fracture) is thus likely to occur, because quick cooling and variations in expansibility create excessive thermal stresses in the weldment. Annealing softening in the HAZ and SC at the root concavity further reduce the strength of the weldment. DMWs therefore tend to follow the regular HAZ fracture mode, with faults originating in the root concavity and propagating along cracks, pores, or the HAZ.

Figure 49  Vickers hardness test on weld cross-section with SMW and DMW (the left alloy with low Al content) for AZ-series alloys.
Figure 50  Vickers hardness test on weld cross-section with SMW and DMW (left alloy is pure Mg) for Mg-Al-based alloy.