Hierarchical Morphology and Formation Mechanism of Collision Surface of Al/Steel Dissimilar Lap Joints via Electromagnetic Pulse Welding

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Abstract: The aim of this study was to characterize detailed microstructural changes and bonding characteristics and identify the formation mechanism of collision surface of Al6061–Q355 steel dissimilar welded joints via electromagnetic pulse welding (EMPW). The collision surface was observed to consist of five zones from the center to the outside. The central non-weld zone exhibited a concave and convex morphology. The welding-affected zone mainly included melting features and porous structures, representing a porous joining. The secondary weld zone presented an obvious mechanical joining characterized by shear plateaus with stripes. The primary weld zone, characterized by dimples with cavity features, suggested the formation of diffusion or metallurgical bonding. The impact-affected zone denoted an invalid interfacial bonding due to discontinuous spot impact. During EMPW, the impact energy and pressure affected the changes of normal velocity and tangential velocity, and in turn, influenced the interfacial deformation behavior and bonding characteristics, including the formation of micropores which continued to grow into homogeneous or uneven porous structures via cavitation, surface tension, and depressurization, along with the effect of trapped air.

Keywords: electromagnetic pulse welding; Al/steel dissimilar joint; collision surface; hierarchical morphology and microstructure

1. Introduction

Dissimilar joining of aluminum alloy and steel as a potential technique to manufacture lightweight, cost-effective, and environmentally friendly structures has been increasingly applied in the transportation industry [1–5]. However, it is difficult to obtain sound joints using conventional fusion welding techniques due to different melting temperatures (~660 °C for Al and ~1538 °C for steel), the severe segregation resulting from mutual dissolution at Al/steel interface, and the formation of brittle intermetallic compounds (IMCs), such as FeAl3, Fe2Al5, FeAl2, FeAl, etc. Electromagnetic pulse welding (EMPW), as an innovative high-velocity impact solid-state welding technology, needs no heat input and
instantaneously completes the welding process (within approximately 40–60 µs), which could overcome the segregation and reduce or suppress the generation of IMCs. Hence, it has recently attracted great attention from researchers in the manufacturing field. EMPW is based on the principle of electromagnetic induction, where the interaction of eddy current and magnetic field generates an enormous Lorentz force driving the flying sheet (Al), striking the target sheet (steel) at a high velocity to achieve Al/steel interfacial bonding. In the welding process, the induced impact energy along with a proper collision angle leads to severe plastic deformation at the interface, causing the rupture of contacting surfaces and the material jetting and removal, then resulting in a deformed and rough fresh surface. Eventually, material contacts and interactions extend along the interface at a high velocity, forming a variety of characteristics such as wavy and vortex diffusion layer, local melting and porous structure, etc. [6,7], which significantly affect the interfacial bonding properties of Al/steel joints.

Some studies on the interfacial morphology, microstructure, and bonding mechanism of EMPWed Al/steel joints have been reported. Lu et al. [8] observed effective interfacial bonding areas of EMPWed AA1060/AISI304 pipe joints, indicating that the wave-like interlock microstructure played a crucial role in the effective bonding. However, detailed studies on the microstructure and its effect on the bonding behavior are absent. Ghosh et al. [9] studied the interface and bonding characteristics of a plain carbon steel sheet joint welded at a discharge voltage of 16 kV and an initial gap of 0.8 mm. They observed three distinct bonding zones, i.e., solid-state bonding, homogeneous-liquid-state bonding, and liquid-state bonding with pores and cracks, but the formation mechanisms need to be identified. Pourabbas et al. [10] reported a porous structure existing in the AA7075/AA4014 pipe joint, which might account for the freezing of trapped air or vaporized aluminum during swift solidification. Geng et al. [11] studied the interfacial transition zone of an EMPWed AA5182/HC340LA lap joint and concluded that the Fe element appeared in three forms in the diffusion zone: Fe–Al IMCs, Fe–Al supersaturated solid solution, and dispersed Fe particles in Al matrix. In addition, they also observed that the thinner melting layer underwent an incredibly fast cooling rate to facilitate amorphization, while the thicker melting layer experienced recrystallization to form ultra-fine grains in a transition zone due to the lower cooling rate [12]. Furthermore, Fan et al. [13] illustrated that the mechanical lattice instability and interdiffusion jointly promoted the amorphization at the Al/Fe interface in the ultra-high velocity impact process. However, the effect of fast local melting features/products and their distribution on the bonding performance is still unclear. In another investigation, Li et al. [14] conducted an experimental study and modeling to clarify the interfacial evolution mechanism and thermomechanical kinetics under extremely high strain rate collision via EMPW. Their results revealed the formation mechanism of the wake, vortex, swirling, and mesoscale cavities due to shear instability with increasing impact energy and illustrated that melting and depressurization facilitated the generation of pores under rapid solidification. Despite the above studies on the cross-section of Al/steel joints, the detailed observations on the collision surface are limited. The relationship among impact parameters, collision products, and their distribution and evolution mechanism is unclear. The effect of microstructural characteristics on interfacial bonding performance has not yet been fully understood. In addition, how the hierarchical morphology is defined along with its formation mechanism in the complicated EMPW collision surface has not been reported in the open literature. The present study was, therefore, aimed to analyze the morphologies and microstructures (such as various dimples, shear plateaus, thin melting features with cracks, diversified porous structures, etc.) on the collision surface of EMPWed Al/steel dissimilar lap joints, identify the multi-zones and formation mechanisms of annular welded area, and finally elucidate the influence of impact parameters and collision products on the bonding properties.
2. Materials and Methods

The commercial 6061-O Al alloy sheets and Q355B high-strength low-alloy (HSLA) steel sheets were used in the EMPW tests, with their chemical compositions listed in Tables 1 and 2. The Al and steel sheets of 1 mm in thickness were machined into coupons of dimensions of 20 mm × 100 mm, with the length being parallel to the rolling direction. All the sheet surfaces were cleaned with ethanol, then dried before welding, being free of burr, oil, indentation, scratch, and other defects.

Table 1. Chemical composition (wt.%) of the 6061-O Al alloy.

| Material       | Mg | Si | Cu | Mn | Fe | Zn | Ti | Cr | Ni | Al |
|----------------|----|----|----|----|----|----|----|----|----|----|
| 6061 O-Al alloy| 1.0| 0.7| 0.3| 0.15| 0.2| 0.25| 0.03| 0.15| 0.05| Bal |

Table 2. Chemical composition (wt.%) of the Q355B steel.

| Material | C   | Si   | Mn   | P   | S   | Cr  | Ni  | Cu  | N   | Fe  |
|----------|-----|------|------|-----|-----|-----|-----|-----|-----|-----|
| Q355B    | 0.24| 0.55 | 1.6  | 0.035| 0.035| 0.3 | 0.3 | 0.4 | 0.012| Bal |

A Pulsar electromagnetic pulse welding system of 70 kJ capacity and capacitance of 240 µF (Chongqing Pulsa Technology Co. Ltd., Chongqing, China) along with an E-shaped one-turn flat coil actuator covered by insulator was utilized to perform EMPW of the dissimilar Al/steel samples. According to the rated power and properties of the equipment, the welding parameters of discharge voltage of U = 14 kV and initial gaps of g = 1 mm, 1.2 mm, and 1.5 mm were used.

The Al/steel EMPW process is schematically illustrated in Figure 1a. First, the initial gap between sheets is adjusted via changing the spacer height based on the assembly and fixing. When the charging system completes the charging of the capacitor bank (C) and the discharge switch is closed, the eddy current induced by a huge instantaneous current (I) occurs on the Al sheet. According to the principle of electromagnetic induction, an enormous Lorentz force generated by the interaction of eddy current and magnetic field drives the flying sheet (Al), hitting the target sheet (steel) at a high velocity to achieve Al/steel interfacial bonding. After welding, the EMPWed Al/steel joints have a total length of 160 mm with an overlap of 40 mm (Figure 1b), forming the annular welding zone along with the central non-weld zone.

Tensile lap shear tests were performed on 5 equivalent samples in each condition (with the same U but different initial gaps) using a fully computerized united testing machine at a constant crosshead speed of 1 mm/min at room temperature. Based on the results of interfacial failure, microstructural characteristics on the fracture surfaces (i.e., collision surfaces) were observed using a scanning electron microscope (SEM, JEOL JSM-6610LV, JEOL, Tokyo, Japan) equipped with energy-dispersive X-ray spectroscopy (EDS).
Figure 1. Schematic diagram of electromagnetic pulse welding (EMPW) process of Al–steel dissimilar materials. (a) Principle of EMPW; (b) dimensions and annular weld zone of a tensile lap shear test sample.

3. Results

According to the results of tensile lap shear tests, the samples made in a condition of $U = 14$ kV and $g = 1$ mm fractured in the Al base metal with the failure location far from the weld zone, implying that the bonding strength of the joints is higher than that of the Al base alloy. Meanwhile, the $g = 1.2$ mm and $1.5$ mm samples tested failed at the weld interface in a mode of interfacial failure, indicating that the load-bearing capacity of the joint does not reach that of Al and steel base metals. Additionally, similar morphologies and microstructures on the fracture surface of both gapped samples were observed by SEM. Hence, the $g = 1.2$ mm sample was selected to examine the details of collision surfaces. Figure 2a shows an overall view of annular welding zone on a typical interfacial fracture surface (i.e., collision surface) on the steel side of an Al/steel dissimilar lap joint, where zone A is selected to conduct detailed SEM examinations and analyses since similar microstructures are present along radial directions (R1, R2, R3). As seen from Figure 2b, the magnified view of zone A with an image rotation of $90^\circ$ could be classified into five layers from I to V due to different morphologies and characteristics. Layer I is the outer non-weld zone, layers II, III, and IV belong to the annular welding zone, and layer V represents the inner (or central) non-weld zone. In addition, based on the previous study [6], the outer zone of the annular welding zone mainly involves metallurgical bonding, and the inner zone mainly exhibits mechanical bonding, which demonstrates the consistency of both observations. The microstructural evolution and formation mechanism of each zone will be sequentially discussed.

3.1. Local Impact Characteristics in Layer I

As seen from Figure 3a, layers I and II exhibit some traces of collision and dimple features, respectively, which are separated by an obvious boundary. Figure 3b shows a magnified view of zone A in Figure 3a; the surface is relatively flat with only several discrete features, i.e., the elongated micro-dimples with the shape of one end being open, metal debris with cavities, and pits. The EDS elemental maps of Figure 3b are presented in Figure 3c,d, indicating that Al alloy is present in both dimpled area and debris but not at
the pits. Accordingly, it could be concluded that local, discrete impact occurs in layer I near the border line, caused by the jetted liquid or splash from the annular welding zone during the high-speed impact welding. The micro-shear dimples denote local spot bonding, which might be caused by relatively larger jetted splashes linked to both sheets. The pits are also hit by the jetted splashes, but the sizes are smaller without contact with both sheets, which are also loose and gone.

Figure 2. (a) An overall view of the interfacial fracture surface on the steel side of an EMPWed Al/steel tensile lap shear sample, and (b) a magnified view of zone A (with an image rotation of 90°), showing a hierarchical morphology from layers I to V.

Figure 3. The evolution of morphology and characteristics in (a–d) layer I and (e–l) layer II: (a) a low magnification image showing layer I, layer II, and their boundary; (b) a magnified image of area ‘A’ in (a), showing the impact features in layer I (near the boundary); (c,d) EDS elemental maps of (b); (e) a magnified image of area ‘B’ in (a); (f,g) EDS line scan results along lines 1 and 2 in (e); (h,i) EDS elemental maps of (e); (j–l) typical dimple fracture features in (e).

3.2. Dimple Feature in Layer II

The magnified view of zone B in Figure 3a is shown in Figure 3e, mainly exhibiting dimple features with varying sizes of cavities. Figure 3f,g present the EDS line scan results, where line 1 across the boundary between layers I and II is shown in Figure 3f, and line
2 across the mainly dimpled region in layer II is shown in Figure 3g. It is clear that the dimpled region basically consists of Al with some extent of fluctuations, which sticks on the Fe side. This could be better seen from the EDS elemental maps (Figure 3h,i), suggesting superior interfacial bonding in this layer due to the nature of ductile fracture or tearing apart from the Al side. Additionally, along the arrow direction in Figure 3h,i, increasingly more of the Al layer is stuck on the Fe, exhibiting an interfacial bonding gradient due to the disturbing of trapped air. Guo et al. [15] also reported a similar change trend of welding cavity in the underwater laser welding of 304 stainless steel, which could be optimized by adjusting laser power and welding speed. Figure 3j–l shows some typical images on the elongated dimple features, i.e., large and small dimples, shallow dimples with voids. Again, the shear fracture is the main feature, reflecting typical interfacial ductile fracture due to plastic deformation and tearing during the interfacial failure of well-bonded lap joints, suggesting that diffusion or metallurgical bonding occurs in layer II. Similar features were also reported in [16]. Sarvari et al. [17] observed both cross-section and collision surfaces, with a well-bonded interface featured by a wavy diffusion layer in the outer annular welding zone, which is in good agreement with the present observations.

3.3. Shear Plateau in Layer III

Based on the difference in the morphologies and characteristics, two boundaries could be further defined in Figure 4a, i.e., boundary 1 between layers II and III-1, and boundary 2 between layers III-1 and III-2. Figure 4b presents the enlarged view of zone A in Figure 4a, exhibiting some distinct shear plateaus along with metal debris (yellow dashed curve boxes) in layer III-1. Some shear plateaus also contain micro-dimples, demonstrating that the shear plateau undertakes the shear force in this layer during the tensile lap shear test. EDS elemental maps of Figure 4b are shown in Figure 4c,d, demonstrating that both dimples and shear plateaus are mainly composed of Al while the metal fragments are mixed Al and Fe. The EDS line scan result of line 1 in Figure 4b is plotted in Figure 4f, and it appears that the shear plateau contains some dispersed Fe, unlike the dimple zone as discussed in Section 3.2. This suggests that mechanical joining was present in layer III-1, and the tensile lap shear test led to the scrubbed shear plateaus. This is in agreement with our previous observation on the interfacial bonding of EMPWed Al/Cu joint, where the irregular or non-uniform mechanical interlocking occurred adjacent to the diffusion bonding zone [6], as discussed in Section 3.2.

Figure 4e shows a magnified view of zone B in Figure 4a, where more local melting areas and porous structures are observed, in addition to some smaller discrete shear plateaus. As seen from the EDS line scan result (Figure 4g) of line 2 in Figure 4e, a heterogeneous composition distribution is present due to instantaneous melting and highly non-equilibrium solidification during EMPW, where Fe concentration is higher in both melting areas and porous structures. This can be seen from Figure 4h being a further magnified view of zone C in Figure 4e, as indicated by red arrows (local melting) and blue dashed circles (cracks). The further magnified view of zone D in Figure 4h is shown in Figure 4i, displaying a distinct T-shaped crack. In the process of EMPW, a peak temperature up to 2750 °C could be reached [18] at a rapid heating rate of ~10⁹ °C/s [19], followed by an ultra-fast cooling rate of ~10¹⁰ °C/s [20]. Such an instant local melting followed by swift solidification would result in large residual stresses causing microcracks and even thin amorphous layers. Wu and Shang [21] also reported that microcracks formed under the condition of a high temperature gradient, a high thermal stress, and a large plastic deformation. Meanwhile, the shrinkage pores and microcracks are not only the supporting evidence for the formation of liquid, but also regarded as a clear indication of rapid melting and solidification at the interface in the EMPW of steel sheets [9]. Hence, the occurrence of amorphous layer is not surprising, which has been confirmed by TEM in the EMPW of Al-Fe bimetallic systems [13,22].
3.4. Local Melting and Porous Structure in Layer IV

Figure 5a shows a low-magnification image exhibiting a relatively legible contrast between layers III and IV, where layer IV is mostly covered by Al as also revealed by ESD line scan along line 1 across the boundary. However, the Al is non-uniformly present along with Fe (Figure 5e). Figure 5b presents an enlarged view of zone A in Figure 5a, exhibiting the distinguishable melting feature and porous structure along with debris. Figure 5c,d present the EDS elemental maps of Figure 5b. It is apparent that Al-Fe compounds account for most parts of zones, except debris which contains more Al and less Fe. It is interesting to note from Figure 5f that the melting area marked by the yellow dashed box appears to be a thin ‘melting membrane’ covering micropores (approximately 0.06–0.5 μm), somewhat like a wave of metallic fluid experiences instantaneous solidification. The EDS results of points 1 and 2 (in Figure 5f) show a composition (at.%.) of roughly 65.3 Al, 34.7 Fe and 78.2 Al, 21.8 Fe, respectively. Therefore, the ‘melting membrane’ might consist of a FeAl3 intermetallic compound, which would be anticipated by viewing the Fe-Al phase diagram since the eutectic temperature is lower than that for the formation of other Al-
Fe intermetallic compounds, such as Fe$_2$Al$_5$, FeAl$_2$ and FeAl, despite the highly non-equilibrium solidification in the EMPW. Lee et al. [23] used TEM to confirm much finer Al-Fe intermetallic compound particles during magnetic pressure seam welding, but they were too small to be identified via electron diffraction pattern. Other intermetallic compounds, e.g., Fe$_4$Al$_{13}$, FeAl and Al$_8$Fe$_{14}$ were reported in the vaporizing foil actuator welding [24]. In addition, Geng et al. [11] also observed FeAl$_2$ in the intermediate layer during EMPW. Figure 5g presents an enlarged view of zone B in Figure 5a, exhibiting more dispersed debris, and the EDS analysis of point 3 presents a composition (at.%) of roughly 92.1 Al, 7.9 Fe. Hence, local melting and its subsequent splashing occurred in layer IV. This is similar to our previous observations in an electromagnetic pulse welded Al/Cu dissimilar joint [6], where a great number of discontinuous splashes in the form of ‘droplet-like’ features or particles with a mixture of Cu and Al even combined with oxygen are dispersed between the inner side of annular welding zone and non-weld zone.

![Figure 5](image_url)

**Figure 5.** The evolution of the melting feature and porous structure in layer IV: (a) A low-magnification SEM image showing layers III and IV; (b) a magnified view of area ‘A’ in (a), showing the melting feature and hierarchical porous structure; (c,d) EDS elemental maps of (b); (e) EDS result along line 1 in (a); (f) a further magnified image of area ‘C’ in (b), exhibiting an architecture of thin membrane covering micropores; (g) a magnified image of area ‘B’ in (a), showing the debris features.

### 3.5. Concave and Convex Morphology in Layer V

Figure 6a demonstrates a change from layers IV to V, denoting an overall impact deformation. From the EDS elemental maps shown in Figure 6b,c, layer V with mostly Fe apparently belongs to the inner non-weld zone. A magnified view of a certain location in layer V is shown in Figure 6d, where a white splashed droplet-like particle is further magnified in Figure 6e. The EDS analysis of point 4 (48.3 Fe, 51.7 O) reveals the occurrence of Fe–O compounds, which were formed due to the interaction of iron and enclosed air in the central non-weld zone at high temperatures. Bellmann et al. [18] performed the calculation and simulation of EMPW of EN AW-6060/C45 steel pipe joint and reported an instant temperature of steel surface that could reach up to 2750 °C, which is far beyond the melting point of iron (1538 °C), leading to the local melting and splashes with some iron oxides. Thus, layer V exhibits a clear concave and convex morphology in Figure 6a, along with splashed droplets in the central non-weld zone.
4. Discussion

4.1. Microstructural Evolution and Formation Mechanism

The above-detailed observations on the collision surface or interfacial fracture surface can be summarized in Figure 7. During EMPW, the strong electromagnetic force (Lorentz force) first drives the Al sheet, hitting the steel sheet at a high velocity and high pressure and generating a high temperature at the interface; then, the impact energy and pressure decrease while the collision angle increases (i.e., $\theta_3 < \theta_2 < \theta_1$ in Figure 7). This rapid process eventually leads to the formation of a special collision surface with multiple zones, which are defined as impact-affected zone (I), primary weld zone (II), secondary weld zone (III), welding-affected zone (IV), and non-weld zone (V) from outside to the center. The initial collision occurs in the central zone or non-weld zone (in Figure 7) and subsequently advances to the outside. The collision velocity $V_c$ could be divided into the normal component $V_n$ (perpendicular to the interface) and tangential component $V_t$ (parallel to the interface), as illustrated in Table 3. The $V_n$ causes severe plastic deformation due to high-velocity impact, forming a concave zone accompanied by an interfacial springback phenomenon; the $V_t$ promotes the outward movement of the collision/shock waves while expels the deformed and extruded material to form a convex zone. However, even for very high impact energy, the co-effect of $V_n$ and $V_t$ is insufficient for jetting to achieve interfacial bonding in the central non-weld zone because the interfacial metallurgical bonding could occur only when an effective jet is formed under a certain collision angle with sufficient impact energy during EMPW. The collision angle in the central non-weld zone is smaller than a critical angle to suppress the jet formation. Moreover, the interfacial springback effect induced by $V_n$ leads to a reflected shock wave, which further restrains the central interfacial bonding or even separates the interface. Geng et al. [25], Wang et al. [26], and Chen et al. [24] also observed and explained this central non-weld zone in Al–Fe magnetic pulse welding, Al–Cu laser impact welding, and Al–steel vaporizing foil actuator welding, respectively.
Figure 7. The evolution of hierarchical morphology of collision surface with changing impact energy, collision angle, and depressurization.

Additionally, several attempts have been reported to reduce or even eliminate the central non-weld zone, e.g., via adjusting the initial collision angle [27,28] (to form an oblique impact between sheets), coil design [29] (to alter the magnitude and distribution of magnetic fields), overlap length [11] (to transform the impacting mode from ‘double-orientation’ to ‘single-orientation’ by shortening it), or adding an Al driver sheet [30] (to reduce negative pressure acting at the collision center). More effective methods in reducing the central non-weld zone and enhancing the bonding area should be further explored in future studies.

As the energy dissipates and the collision angle increases, the tremendous quantity of heat arising from severe plastic deformation, Joule heating, collision energy dissipation, and supersonic gaseous compression [20,31,32] give rise to the local melting in the welding-affected zone or layer IV, accompanied by porous structures and splashed metal debris, as shown in Figure 5. In this stage, the surface tension of melt, dissolved or trapped air, instant high temperature, and pressure provoked by a high-velocity impact (being as high as 1000 m/s [23]) would affect the formation of porous structure and the further interfacial bonding. Qu et al. [33] reported that the heterogeneous porous structure could be interpreted as a result of dissolved air in the molten metal; thereby, the trapped air nucleates to form micropores at the high velocity and pressure, then gradually grows into the characteristic porous structure with the rapid cooling (~10¹⁰ °C/s [20]) of the molten mass. Rice and Tracey established a void growth model, where the rate of growth depends on the hydrostatic stress, von Mises norm of the Cauchy stress tensor (the equivalent stress), and equivalent plastic strain [34,35]. Therefore, one could deduce that the porous structure in the EMPW process may be mainly attributed to the successive nucleation of micropores and the growth of spherical pores during the intense plastic deformation under rapid heating and cooling with instantaneous high depressurization at the interface. First, the cavitation phenomenon happens to maintain the conservation of the total volume in the material subjected to an instantaneous local melting due to an ultra-high heating rate of ~10⁹ °C/s [19], followed by the rupture of melt due to a significantly large inherent surface
tension, leading to the nucleation of micropores. Then, the interface undergoes a sudden decrease of temperature at a rate of as high as ~10^{10} °C/s, accompanied by an ultra-high depressurization rate in the order of ~10^{11} MPa/s [20]. The rapid expansion induced by such an interfacial depressurization promotes the radial growth of each micropore to form a spherical pore, and the solidification shrinkage also continuously contributes to the coalescence of pores, resulting in the formation of a regionalized porous structure, as schematically shown in Table 3. In addition, a similar thin melting membrane porous structure was also reported in [20], which could be an amorphous structure without grain boundaries generated under high strain, local heating, and rapid solidification. Further studies would be to explore how the porous structures are formed via vacuum electromagnetic pulse welding, where the interference of air could be precluded.

Table 3. Schematic diagrams showing the effect of impact-related factors on the interfacial characteristics.

| V (Non-Weld Zone) | IV (Welding Affected Zone) | III (Secondary Weld Zone) | II (Primary Weld Zone) | I (Impact Affected Zone) |
|-------------------|---------------------------|---------------------------|------------------------|-------------------------|
| Impact energy     | $E_5$                     | $E_4$                     | $E_3$                  | $E_2$                   | $E_1$                   |
| Collision angle   | $\theta_5$                | $\theta_4$                | $\theta_3$             | $\theta_2$              | $\theta_1$              |
| Impact velocity   | $V_{i5}$                  | $V_{i4}$                  | $V_{i3}$               | $V_{i2}$                | $V_{i1}$                |
| Pressure          | $P_5$                     | $P_4$                     | $P_3$                  | $P_2$                   | $P_1$                   |

![Diagram showing the effect of impact-related factors on the interfacial characteristics](image)

| Bonding state     | Unbonded                  | Porous joining            | Mechanical joining    | Diffusion or metallurgical bonding | Spot joining |
|-------------------|---------------------------|---------------------------|-----------------------|-----------------------------------|--------------|
| Fracture location | Non                       | Less                       | Less                  | Less                              | Less         |
The secondary weld zone or layer III in Figure 4 consists of layer III-1 (large areas of shear plateaus along with metal debris) and layer III-2 (small and scattered shear plateaus with local melting and porous structure). This reflects a significant influence of welding parameters in Table 3 on the impact behavior and jetting, eventually leading to essential changes in the interfacial bonding performance, i.e., layer III-1 has a superior load-bearing capacity to that of layer III-2. Additionally, the intense plastic deformation at a strain rate of up to $10^6$–$10^7$ s$^{-1}$ [36] and interfacial shear instability leads to the instant rupture and ejection of softened or liquefied metal, then forming the debris with varying sizes and shapes (Figure 4b,e and Figure 5b,g).

The primary weld zone or layer II is mainly characterized by dimples along with cavities, as shown in Figure 3e. In this layer, although the collision energy has been partially dissipated, the collision angle has reached an optimal range to generate jetting (as shown in Table 3), which prompts exposure of fresh surfaces and then contact of each other to form diffusion or metallurgical bonding. While a small amount of air is captured between sheets due to the pressure decrease, the depressurization facilitates the trapped air to generate cavities of various sizes (as indicated in Table 3), leading to the morphology of dimples surrounding large and small cavities. In general, dimple features reflect the occurrence of plastic deformation and ductile fracture, indicating a well-bonded interface. As cavities cause stress concentration and premature interfacial failure, they should be reduced or avoided. However, as noted in Figure 3a,e,h,i, the cavity is not innate but gradually generated from none to some under varying collision energy and collision angle, i.e., it starts from the lower to the upper boundary within layer II. Therefore, it could be inferred that if the discharge voltage or energy ($U$ or $Q$, affecting collision energy) and the initial gap ($g$, affecting collision angle) are altered to appropriate values, the cavities could be eliminated or reduced due to the expelling of trapped air. The overall continuous bonding was observed in the EMPW of Al–Fe joints welded at $Q = 40$ kJ and $g = 1.4$ mm [37], Al–Ti joints made with $Q = 35$ kJ and $g = 1.4$ mm [38], and Al–Cu joint welded at $U = 12$ kV and $g = 1.0$ mm [6]. As a result, despite some extent of cavities existing in the dimple zone, continuous and superior interfacial bonding is present in layer II, thus being the primary weld zone.

In the final stage of EMPW, the massive dissipation of collision energy results in an extreme decrease in pressure so that a large amount of air that cannot be driven out instantly is freely compressed in-between the sheets; thus, the non-uniform collision happens in the impact-affected zone or layer I. This results in the interfacial spot bonding at some sites of certain energy concentration, presenting the micro-dimple features after tensile lap shear tests, along with some pits observed in Figure 3b.

**4.2. Effect of Impact Parameters on Interfacial Characteristics**

From the above formation mechanisms discussed in Section 4.1, one could conclude that the impact (or welding) parameters, including collision energy and angle, impact velocity, and pressure, affect the changes of fracture morphologies and characteristics. As summarized in Table 3, during EMPW, both impact energy and impact velocity across five layers from the inside to the outside gradually decrease ($E_5 > E_4 > E_3 > E_2 > E_1$; $V_{c5} > V_{c4} > V_{c3} > V_{c2} > V_{c1}$), while collision angle gradually increases ($\theta_5 < \theta_4 < \theta_3 < \theta_2 < \theta_1$). Then, the normal component $V_n (V_n = V_c \times \cos \theta)$ also decreases, while the tangential component $V_t (V_t = V_c \times \sin \theta)$ potentially increases first and then decreases, accompanied by the reduction of pressure across five layers ($P_5 > P_4 > P_3 > P_2 > P_1$). Indeed, $V_{c5}$ and $V_t$ are two key variables influencing the periodic interfacial interaction throughout the overall impact process. In the non-weld zone or layer V, $V_{c5}$ keeps the maximum while $V_{c5}$ is small; the pressure wave (generated by $V_{c5}$) along with the surface wave (induced by $V_{c5}$) results in a concave and convex morphology in the central non-weld zone (Figure 6a).

With the dissipation of energy in the welding affected zone (layer IV), $V_{n4}$ decreases while $V_{t4}$ increases, which weakens the springback effect but prompts the removal of materials to generate jetting. Meanwhile, the reflection wave starts to encounter the collision point,
which results in local melting and porous features in the welding-affected zone. As the collision continues, \( V_n \) decreases in the secondary weld zone (layer III) and primary weld zone (layer II), while \( V_t \) increases first and then decreases, with layer III being wider than layer II. More importantly, the collision angle reaches a reasonable or optimal range within these two layers, where the cooperation between \( V_{n3} \) and \( V_{t3} \) or \( V_{n2} \) and \( V_{t2} \) leads to mechanical joining in the secondary weld zone (Figure 4) and diffusion (or metallurgical) bonding in the primary weld zone (Figure 3), as also summarized in Table 3. In the impact-affected zone (layer I), both \( V_{n5} \) and \( V_{t5} \) decrease to the minimum, coupled with the large depressurization and air compression, rarely causing an interfacial impact. Sapanathan et al. [39] also reported that normal and tangential velocities stemming from the collision angle influenced the interfacial morphology of EMPWed Al/Al and Al/Cu pipe joints.

Based on the stress (or pressure) wave theory, pressure wave and surface wave will occur after the collision. Yu et al. [40] illustrated the formation mechanism of a wavy interface in an EMPWed Al/Al pipe joint. The wavy characteristics appeared at the initial stage of impact; however, the velocity of the reflection wave is lower than that of the compression wave, so that the reflection wave acts on the interface behind the compression wave, which hinders interfacial bonding or even separates the interface. This is in agreement with the present result (Figure 2). Ben-Artzy et al. [41] illustrated the formation process of interfacial waves in detail based on a Kelvin–Helmholtz instability mechanism, where the periodic variation of compression and reflection waves in the vicinity of collision point leads to the interfacial interaction. Gradually, the compression wave from the central collision point to outside slows down due to the dissipation of impact energy; when the reflection wave catches up and interacts with the compression wave, the inter-friction leads to high temperatures at the interface and highly accelerated non-equilibrium inter-diffusion, allowing for the interfacial bonding. In addition, as shown in Figures 3–5, the change of impact energy \( E \) and pressure \( P \) results in a regional difference of cavity or porous characteristics. As summarized in Table 3, \( E \) and \( P \) are larger in layer IV, cavitation effect and melt surface tension along with some trapped air prompt the nucleation of micropores, and depressurization further promotes the radial growth of micropores after tensile lap shear tests. Furthermore, the fracture locations also reflect different bonding performances, i.e., the welding-affected zone (layer IV) presents a relatively inferior porous joining, and interfacial failure occurs in the porous structure; the secondary weld zone (layer III) shows a mechanical joining, and the interfacial fracture mainly appears at the site close to Al side and occasionally tears in the Al matrix; the primary weld zone (layer II) exhibits a diffusion or metallurgical bonding, and joint failure always locates on the Al side, as schematically illustrated in Table 3.

5. Conclusions

The main conclusions of this study can be summarized as follows:

(1) During EMPW, a strong electromagnetic force propelled the Al sheet, hitting the steel sheet at a high velocity and creating a robust dissimilar lap joint with a special collision surface containing multi-zones. Based on the interfacial microstructure and bonding state, the collision surface was defined as the central non-weld zone, welding-affected zone, secondary weld zone, primary weld zone, and impact-affected zone from the center to outside.

(2) The central non-weld zone exhibited a concave and convex morphology. The welding-affected zone mainly included some melting features and porous structures, representing a porous joining. The secondary weld zone presented an obvious mechanical joining characterized by the shear plateaus with stripes, and the instantaneous melting and rapid solidification features were prone to cracking due to the presence of residual stresses. The primary weld zone, characterized by dimples with cavity features, indicated the formation of diffusion or metallurgical bonding. The impact-affected zone denoted an invalid interfacial bonding due to the discontinuous spot impact.
(3) The impact energy and pressure influenced the changes of normal velocity and tangential velocity, and in turn, affected the interfacial deformation behavior and bonding characteristics, including the formation of micropores, which continued to grow into homogeneous or uneven porous structures under cavitation, surface tension, and depressurization, along with the effect of trapped air.

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References
1. Xu, Z.D.; Cui, J.J.; Yu, H.P.; Li, C.F. Research on the impact velocity of magnetic impulse welding of pipe fitting. Mater. Des. 2013, 49, 736–745. [CrossRef]
2. Chang, S.Y.; Tsao, L.C.; Lei, Y.H.; Mao, S.M.; Huang, C.H. Brazing of 6061 aluminum alloy/Ti-6Al-4V using Al-Si-Cu-Ge filler metals. J. Mater. Process. Technol. 2012, 212, 8–14. [CrossRef]
3. Kazantseva, N.; Krakhmalev, P.; Thuvander, M.; Yadroitsev, I.; Vinogradova, N.; Ezhov, I. Martensitic transformations in Ti-6Al-4V (ELI) alloy manufactured by 3D Printing. Mater. Charact. 2018, 146, 101–112. [CrossRef]
4. Wei, S.Z.; Li, Y.J.; Wang, J.; Liu, K. Formation of brittle phases during pulsed current gas tungsten arc welding of titanium to aluminum alloys. J. Mater. Eng. Perform. 2014, 23, 1451–1457. [CrossRef]
5. Tarui, T. State of the art of joining technology for light weight in automotive industry. In Proceedings of the Mechanical Engineers Congress Japan 2016, Tokyo, Japan, 11–14 September 2016.
6. Wang, P.Q.; Chen, D.L.; Ran, Y.; Yan, Y.Q.; She, X.W.; Peng, H.; Jiang, X.Q. Electromagnetic pulse welding of Al/Cu dissimilar materials: Microstructure and tensile properties. Mater. Sci. Eng. A 2020, 792, 139842. [CrossRef]
7. Wang, P.Q.; Chen, D.L.; Ran, Y.; Yan, Y.Q.; Peng, H.; Jiang, X.Q. Fracture characteristics and analysis in dissimilar Cu-Al alloy joints formed via electromagnetic pulse welding. Materials 2019, 12, 3368. [CrossRef]
8. Lu, Z.Y.; Gong, W.T.; Chen, S.J.; Yuan, T.; Kan, C.L.; Jiang, X.Q. Interfacial microstructure and local bonding strength of magnetic pulse welding joint between commercially pure aluminum 1060 and AISI 304 stainless steel. J. Manuf. Process. 2019, 46, 59–66. [CrossRef]
9. Ghosh, P.; Patra, S.; Chatterjee, S.; Shome, M. Microstructural evaluation of magnetic pulse welded plain carbon steel sheets. J. Mater. Process. Technol. 2018, 254, 25–37. [CrossRef]
10. Pourabbas, M.; Abdollah-Zadeh, A.; Sarvari, M.; Pouranvari, M.; Miresmaeili, R. Investigation of structural and mechanical properties of magnetic pulse welded dissimilar aluminum alloys. J. Manuf. Process. 2019, 37, 292–304. [CrossRef]
11. Geng, H.H.; Xia, Z.H.; Zhang, X.; Li, G.Y.; Cui, J.J. Microstructures and mechanical properties of the welded AA5182/HC340LA joint by magnetic pulse welding. Mater. Charact. 2018, 138, 229–237. [CrossRef]
12. Geng, H.H.; Mao, J.Q.; Zhang, X.; Li, G.Y.; Cui, J.J. Formation mechanism of transition zone and amorphous structure in magnetic pulse welded Al-Fe joint. Mater. Lett. 2019, 245, 151–154. [CrossRef]
13. Fan, Z.S.; Yu, H.P.; Li, C.F. Interface and grain-boundary amorphization in the Al/Fe bimetallic system during pulsed-magnetic-driven impact. Scr. Mater. 2016, 110, 14–18. [CrossRef]
14. Li, J.S.; Raelolison, R.N.; Sapanthan, T.; Hou, Y.L.; Rachik, M. Interface evolution during magnetic pulse welding under extremely high strain rate collision: Mechanisms, thermomechanical kinetics and consequences. Acta Mater. 2020, 195, 404–415. [CrossRef]
15. Guo, N.; Fu, Y.L.; Xing, X.; Liu, Y.K.; Zhao, S.X.; Feng, J.C. Underwater local dry cavity laser welding of 304 stainless steel. *J. Mater. Process. Technol.* 2018, 260, 146–155. [CrossRef]
16. Zhang, H.; Jiao, K.X.; Zhang, J.L.; Liu, J. Microstructure and mechanical properties investigations of copper-steel composite fabricated by explosive welding. *Mater. Sci. Eng. A* 2018, 731, 278–287. [CrossRef]
17. Sarvari, M.; Abdollah-Zadeh, A.; Naffakh-Moosavy, H.; Rahimi, A.; Parsaeyan, H. Investigation of collision surfaces and weld interface in magnetic pulse welding of dissimilar Al/Cu sheets. *J. Manuf. Process.* 2019, 45, 356–367. [CrossRef]
18. Bellmann, J.; Lueg-Althoff, J.; Schulze, S.; Hahn, M.; Gies, S.; Beyer, E.; Tekkaya, A.E. Thermal effects in dissimilar magnetic pulse welding. *Matsals* 2019, 9, 348. [CrossRef]
19. Sapanathan, T.; Raoelison, R.N.; Buiron, N.; Rachik, M. In situ metallic porous structure formation due to ultra high heating and cooling rates during an electromagnetic pulse welding. *Scr. Mater.* 2017, 128, 10–13. [CrossRef]
20. Raoelison, R.N.; Li, J.S.; Sapanathan, T.; Padayodi, E.; Buiron, N.; Racine, D.; Zhang, Z.; Marceau, D.; Rachik, M. A new nature of microporous architecture with hierarchical porosity and membrane template via high strain rate collision. *Materialia* 2019, 5, 100205. [CrossRef]
21. Wu, X.; Shang, J.H. An investigation of magnetic pulse welding of Al/Cu and interface characterization. *J. Manuf. Sci. Eng. Trans. ASME* 2014, 136, 051002. [CrossRef]
22. Li, J.J.; Yu, Q.; Zhang, Z.J.; Xu, W.; Sun, X. Formation mechanism for the nanoscale amorphous interface in pulse-welded Al/Fe bimetallic systems. *Appl. Phys. Lett.* 2016, 108, 201606. [CrossRef]
23. Lee, K.J.; Kumai, S.; Arai, T.; Aizawa, T. Interfacial microstructure and strength of steel/aluminum alloy lap joint fabricated by magnetic pressure seam welding. *Mater. Sci. Eng. A* 2007, 471, 95–101. [CrossRef]
24. Chen, S.H.; Daehn, G.S.; Vivek, A.; Liu, B.; Hansen, S.R.; Huang, J.H.; Lin, S.B. Interfacial microstructures and mechanical property of vaporizing foil actuator welding of aluminum alloy to steel. *Mater. Sci. Eng. A* 2016, 659, 12–21. [CrossRef]
25. Geng, H.H.; Mao, J.Q.; Zhang, X.; Li, G.Y.; Cui, J.J. Strain rate sensitivity of Al-Fe magnetic pulse welds. *J. Mater. Process. Technol.* 2018, 262, 1–10. [CrossRef]
26. Wang, X.; Shao, M.; Jin, H.; Tang, H.; Liu, H.X. Laser impact welding of aluminum to brass. *J. Mater. Process. Technol.* 2019, 269, 190–199. [CrossRef]
27. Wang, H.M.; Liu, D.J.; Lippold, J.C.; Daehn, G.S. Laser impact welding for joining similar and dissimilar metal combinations with various target configurations. *J. Mater. Process. Technol.* 2020, 278, 116498. [CrossRef]
28. Lueg-Althoff, J.; Bellmann, J.; Hahn, M.; Schulze, S.; Gies, S.; Tekkaya, A.E.; Beyer, E. Joining dissimilar thin-walled tubes by magnetic pulse welding. *J. Mater. Process. Technol.* 2020, 279, 116562. [CrossRef]
29. Zhang, H.Q.; Yang, K.; Chernikov, D.; Sapanathan, T.; Padayodi, E.; Buiron, N.; Racine, D.; Zhang, Z.; Marceau, D.; Rachik, M. A new nature of microporous architecture with hierarchical porosity and membrane template via high strain rate collision. *Materialia* 2019, 5, 100205. [CrossRef]