Microstructure, Welding Mechanism, and Failure of Al/Cu Ultrasonic Welds

Xin Wu¹, Teng Liu^{1,2} and Wayne Cai³

¹Wayne State University, Detroit, MI, USA

²Shenyang Aerospace University, Shenyang, China

³GM Global R&D Center, Warren, MI, USA

xwu@eng.wayne.edu, tengliu7@gmail.com, wayne.cai@gm.com

Abstract

Ultrasonic metal welding has been used widely to join battery cell terminals, or tabs (either Al or Cu), with bus bars (Cu) to form assembled battery packs in battery electric vehicles. However, the mechanism of ultrasonic welding for Al/Cu is still not well understood. In this work, the microstructures of the ultrasonic welds between three layers of lithium-ion battery tabs (either Al or Cu) and bus bars were studied. From the microstructure analysis, the weld formation mechanism and failure modes were investigated. It was found that the metal inter-mix is the main weld formation mechanism among Al tabs, while diffusion bonding is the main mechanism for Cu-Cu or Al-Cu. Metallographs also indicated that the weld failure is a combination of the interfacial debonding between the innermost tab (either Cu or Al) and the Cu bus bar and the Mode III through-thickness fracture of the tabs. The presented work provides better understanding of weld formation and failure, and hence provide insight into ultrasonic welding process towards weld quality improvement. This understanding and insight can be used to develop science-based design guidelines towards selecting the most appropriate materials (including heat treatment and coating), and welding configurations (such as layers of tabs), and welding process parameters.

Keywords: Ultrasonic welding, welding mechanism, failure, microstructure, aluminum, copper

1 Introduction

In ultrasonic metal welding, a high frequency shear oscillation generated by a piezoelectric system removes surface oxides or contamination by friction. The reciprocating sliding motion under pressure results in metallurgical adhesion at the metal contact interfaces without melting. Such a solid-state bonding is advantageous for joining dissimilar metals such as copper and aluminum, commonly used materials for lithium-ion batteries. Therefore, in recent years, there are tremendous interest in ultrasonic welding of multi-layered battery cell terminals, or tabs, for electrical vehicles. An example of ultrasonic battery tab joining is schematically illustrated in Fig. 1. In this configuration, three layers

of 0.2 mm thick Ni-plated pure copper (Cu) tab or 0.2 mm thick anodized aluminum (Al) tab, and one layer of 0.9 mm thick Ni-plated pure Cu bus bar are welded in a “cccC” or “aaaC” stacking configuration, where “C” refers to the 0.9 mm Cu bus bar, and “c” and “a” refer to the 0.2 mm Cu and Al tabs, respectively.

Figure 1. Three battery cells grouped in parallel and three groups in series by joining tabs (“aaa” or “ccc”) to “U”-shaped Cu bus bars (“C”) as shown in front view (a) and top view (b)

The performance requirements for the battery tab welds are: mechanical strength (against static load, vibration, crash impact, and fatigue), electrical resistance (with low electrical resistivity, low joule heating, and low energy loss), and chemical stability (against moisture and corrosion). It is notable that these requirements can conflict with each other. For example, almost all material strengthening mechanisms, such as alloying, grain refinement, cold work, phase-transformation, or second-phase reinforcement, will increase electrical resistivity as well as corrosion reactivity. The quality and performance of a battery pack highly depends on the soundness of the battery tab joining.

A general description of ultrasonic welding can be found in many welding textbooks or handbooks, for example by Potente (1984), Staff (2008), and Singh (2011). Ultrasonic devices were initially invented for nondestructive evaluation, cleaning and degreasing during the World War II, and later was used in welding of thermoplastics since 1960's. For polymers, a weld is commonly established at temperatures above glass transition temperature T_g . Upon cooling, the weld is generally stronger than the base materials (Matsuoka 1995; Takeda et al. 1998). The recent advancement of the power ultrasonic technology has been a joint effort by academia and industry. For example, researchers published a series of papers on complex ultrasonic weld tips (Tsujuno et al. 1989), two-vibration-system development for multi-spot continuous welding metal plates (Tsujino & Ueoka, 1996), transverse and torsional complex system with more than one single frequency (Tsujino et al. 1998; Tsujino et al. 2000; Tsujino et al. 2004), the use of large weld tip (Tsujino & Sano et al. 2002), and the development of high powered, stepped transverse vibration rods. They also reported research results on high frequency wire bonding (Tsujino & Hasegawa, 1996), metals/plastics welding (Tsujino et al. 1996), aluminum alloys and stainless steel plates welding (Tsujino & Hidai et al. 2002), coated copper wires welding (Tsujino et al. 2004), and Al/Cu welding (Matsuoka & Imai, 2009). Metals/ceramics ultrasonic joining was studied by Matsuoka (1994) and Matsuoka (1998). Al/Al welding was studied by Janaki Ram et al. (2006). The ultrasonic welding deformation was studied by Bakavos and Prangnell (2010) with the use of EBSD. They described the Al/Al ultrasonic welding

mechanism as a series of progressive processes as micro-welding, expansion, convoluted wavy interface formation, vortices/swirls and ripples, thinning and shear band formation. Chen et al. (2012) reported the formation of heat affected zone in ultrasonic spot welding of aluminum AA6011-T4 with an ageing effect. The measured temperature was as high as 400°C which caused softening immediately after welding and then recovered at a rate faster than the parent material. For ultrasonic welding of copper, Elangovan et al. (2010) performed experimental study based on a design of experiment method, and identified several critical process parameters. The ultrasonic welding of nickel coated Cu/Cu tab was studied by Kim et al. (2011). In the study, a factorial experimental design and peeling test were performed, and three categories of weld quality (cold, good and over-welds) was mapped in the welding time-welding pressure domain. For welding dissimilar materials, Bakavos & Prangnell (2010) studied the effect of ultrasonic welding parameters on mechanical and microstructure properties of dissimilar joints of aluminum and stainless steels and the results indicated that the maximum tensile strength of the weld are related to a reasonably amount of bond density and material thinning. The ultrasonic welding process was modeled by Suresh et al. (2007) and Levy et al. (2011) for predicting temperature distribution of polymers using coupled thermomechanical finite element analyses. Other studies for ultrasonic welding of dissimilar metals from different authors include welding mild steel/aluminum-magnesium alloy sheets by Watanabe et al. (2009) and Al/Mg by Panteli et al. (2012).

The ultrasonic welding process may also be used for additive manufacturing from aluminum sheets, as reported by Watanabe & Mori (1996), and Janaki Ram et al. (2006). In the latter case, the substrate was heated to increase the welding effectiveness.

For the present interest of battery tab ultrasonic welding, welding quality not only affects the functionality and energy consumption, but also the durability and reliability of a vehicle. The electrical resistance of the welded tabs typically increases as the welds accumulate micro-damage under a static or fatigue loading when the vehicles are in use. The correlation between fatigue damage and resistance can be recognized from the fact that many researchers have been using this phenomenon as one of the non-destructive testing methods for damage monitoring, as reviewed by Ditchburn et al. (1996). The resistance measurement was used for weld quality inspection by Balle et al. (2011) for testing ultrasonically welded aluminum/carbon-reinforced polymer, by Vavouliotis et al. (2011) for a carbon nanotube reinforced polymer, by Omarri & Sevostianov (2013) for stainless steel, and recently by Sun & Guo (2004).

A review article focused on automotive joining techniques is seen by Barnes and Pashby (2000), who listed the advantages and limitations of the selected joining techniques. Another recent review article by Mori et al. (2013), also focused on automotive application, summarized the deformation-based joining techniques including some traditional and non-traditional material forming processes (e.g. rolling, forging, extrusion, hydroforming, electromagnetic forming and incremental forming, impulse welding or electromagnetic welding), friction welding and friction stir welding, and many direct mechanical fastening methods involving plastic deformation (such as self-pierce riveting, mechanical clinching, hemming, seaming and staking).

Currently, research on lithium-ion battery ultrasonic welding has been very active. The work by Kim et al. (2011) studied the ultrasonic welding for battery tabs with like and dissimilar metals. In a follow-up study by Lee SS, et al. (2013), the quality of ultrasonically welded battery tab was characterized. Zhao et al. (2013) fabricated a thin-film thermocouple inserted in the ultrasonic tools 100 microns away from the tool/workpiece interface. Such measured temperature rise during ultrasonic welding of Ni-coated Cu was as high as 650°C for Cu/Cu welding. With this technique, Li, et al. (2013) studied the transient thermal process during ultrasonic welding of battery tabs that helps better understand the welding mechanism and resultant microstructure and properties. The high temperature, along with the earlier mentioned high strain rate deformation studied by Panteli et al. (2012) on Al/Mg USW, results in a special microstructure evolution, including dynamic recrystallization in Cu/Cu ultrasonic welding reported by Lee SS, et al. (2013). The welding process

was further modeled by Zhao et al. (2014) to predict fatigue life of ultrasonically welded Li-ion battery tabs, and by Lee D., et al. (2013), who simulated the ultrasonic welding process of multiple thin and dissimilar metals. In addition, the dynamics, stress and energy transmission in battery tab ultrasonic welding were studied by (Kang et al. 2013; Lee et al. 2014; Kang et al. 2014(1); Kang et al. 2014(2)).

This paper focuses on weld formation and failure analyses for ultrasonic welding of multi-layered battery tabs involving a thick copper to three thin Cu or Al tabs, by means of metallographic analyses. The experimental condition is given in Section 2. Sections 3 and 4 study the weld formation and failure mechanisms of Cu and Al tabs, respectively, followed by Conclusions in Section 5.

2 Ultrasonic Welding and Shear Test Procedures

Ultrasonic welds were produced from three layers of either Cu or Al tabs and one layer of Cu bus bar (coded as “cccC” and “aaaC”, respectively), using a Stapla ultrasonic welder. Each tab is of 45 mm (length) x 41 mm (width) x 0.2 mm (thickness), and each Cu bus bar is a flat coupon of 45 mm (length) x 41 mm (width) x 0.9 mm (thickness), as shown in Fig. 2. Note that the flat Cu coupons were used instead of the U-shaped Cu bus bar as in battery pack production (Fig. 1) for convenience and cost reasons. Both the Cu tabs and Cu bus bars were electro-plated with nickel (Ni). The Ni-coating layer thickness is about 0.75 micron for Cu tabs and 2 microns for the Cu bus bars. The 0.2 mm pure Al tabs were anodized, with about 0.2 micron of oxidation film on the tab surfaces.

As shown in Fig. 2, each ultrasonic weld consists of three weld spots with a center-to-center spacing of 12 mm. Each weld spot in cccC is approximately 5 mm x 3 mm, and in aaaC approximately 10 mm x 4 mm, both with corner radius of about 1 mm. The three weld spots are collinear along the long axis direction. The ultrasonic vibration direction is along the short axis.

Many parameters can impact the ultrasonic welding quality, ranging from the weld tool design (such as the knurl patterns of the horn, as shown in Fig. 2, and/or anvil) to welding process parameters (e.g., the clamping pressure, the ultrasonic amplitude, and the welding energy). In this study, to reduce the sample size of experiments, the clamping pressure (4.8 bar) and the ultrasonic amplitudes (25 μ m for cccC welding, and 15 μ m for aaaC welding) were held constant. Only welding energies were varied to produce welds at three different quality levels, i.e., under-weld, normal-weld and over-weld. For cccC welding, the three welding energy levels of 1600 J, 2400 J and 3200 J were used to represent the under-weld, normal-weld and over-weld conditions. For aaaC welding, the three energy levels of 300 J, 600 J and 1200 J were used.

The quality of a weld was evaluated by means of lap-shear testing. Since the most critical (and also the weakest) bond for either a cccC or aaaC weld is between the innermost tab and the Cu bus bar (Lee DK., et al. 2013), a lap-shear test configuration in Fig. 2 was used. By clamping the three tabs at one end and the Cu bus bar at another end, the lap-shear tests were performed under a constant speed of 0.5 mm/s on an MTS machine.

Figure 2. (top) Lap shear test setup and specimen dimensions; (bottom) horn knurl pattern on an Al weld

3 Microstructural Analyses and Mechanical Tests of cccC Welds

Optical microscope (OM), Scanning Electron Microscope (SEM), and energy-dispersive X-ray spectroscopy (EDX) were used to characterize the weld quality, fracture surface appearance and chemical composition.

Representative cccC weld images are shown in Fig. 3 at three different welding conditions. The 5 x 3 array of square biting facets were formed due to the horn knurls. Note the different sizes of the biting facets as well as the differences in residues of the Ni-coating (in silver color) on the weld surfaces in three different welding conditions.

Figure 3. Three cccC welds with different levels of welding conditions, i.e., under-weld, normal-weld, and over-weld, as viewed from the Cu tab outside surface.

To better understand the bonding between different layers, a cccC normal-weld was sectioned along the long and short axes of the middle spots, respectively, and the OM micrographs are shown in Fig. 4. As can be seen:

- The Ni-coating at the interfaces, or the bond lines, were continuous and essentially intact. No obvious interfacial separation, cracks, or voids existed. The bond line waviness, though, was relatively significant at the interface of the top two tabs. The waviness then decreased substantially to almost nil at the innermost tab and the bus bar interface.

- For both sections, the material flow and deformation occurred mainly at the top Cu tab with thickness reduced to about a half.

- The long-axis section of the metals have been extruded to form complete “valley” area; however, the metal did not fully fill-in the horn grooved regions to form a complete “Λ”-shaped hill, i.e., a complete tool-workpiece surface contact was not established. On the short-axis section, which corresponds to the ultrasonic vibration direction, a complete tool-workpiece contact for both “valley” (the horn's teeth) and “hill” (the horn's groove) has been established. This is consistent with Fig. 3.

Figure 4. Cross-sections of the middle spot of a cccC normal-weld (2400 J): sectioned along the long axis (top), and short axis (bottom).

The SEM micrographs of the interface between the innermost tab and the Cu bus bar (a.k.a., the ccc|C interface) of the middle spot of a normal-weld is shown in Fig. 5, after detaching all the three tabs from the Cu bus bar. The 5 x 3 teeth impression from the horn was visible, including the contact boundaries. The welded regions showed many long stripes, whose orientation coincided with the ultrasonic shearing directions. Outside the stripe area the surface was relatively flat and smooth, with no sign of interface tearing. This suggests that those stripes areas were strongly bonded, while the remaining areas were not bonded or not strongly bonded.

Figure 5. SEM micrographs of the ccc|C interface viewed on the 0.9 mm Cu bus bar surface, for the central spot of normal-weld specimen and with local enlargements of the area corresponding to the horn knurls at the 2nd row and 4th column.

The EDX chemical analysis by line scanning in the direction perpendicular to the stripe direction is shown in Fig. 6. It can be seen that the Cu base metal was exposed on those striped areas, which further confirms the ductile fracture occurred by peeling out of the base Cu metal or the Ni metal within the coating layer. Since the original Cu sheets were coated with Ni and in the Cu-exposed areas the Ni-Ni bonding is stronger than the base metal itself, the bonding strength is controlled by the Cu strength. In the rest of the weld spot, the fractured surfaces are still covered by the Ni coating layer, which can be either not welded, or weakly welded at lower strength than Ni itself.

Figure 6. The EDX analysis showing the tearing portion with Cu exposed on the fractured surface, while the rest areas (including weld spot area) being still covered by the Ni coating.

Since the weld strength is positively correlated to the total bonding area, it is desirable to determine the actual bonding area or the fraction or percentage of Cu-exposed area versus the nominal total bonding area (a.k.a., the weld spot). Hence, a dark field (DF) microscopy technique was used to observe the metal contents at the ccc|C interface, using an optical microscope, as shown in Fig. 7. The exposed Cu shows in bronze color, and the Ni shows in green. As a comparison, under a bright field (BF) mode, this distinction is not possible. With this technique, the DF images for three different welding conditions are shown in Fig. 7 (bottom). A Matlab program was subsequently developed to calculate the bonding area by identifying the color code associated with each pixel and then marking the pixels within certain color codes as the bonding area. For example, the bronze areas (i.e., the Cu-exposed areas) in Fig. 8 correspond to the Cu areas in Fig. 7.

Figure 7. (top) The comparison of OM bright field (BF) and dark field (DF) images; (bottom) Dark field images at the ccc|C interface: the Cu areas are in bronze and Ni areas are in green, whereas the circled area at the upper-left corner is enlarged and shown on the top.

Figure 8. The bonding area as calculated from the image processing technique.

Based upon the above chemical and image analyses of the fractured surfaces, it is concluded that:

(1) Only a small fraction of the innermost Cu tab (Ni-coated) was bonded with the Cu bus bar (Ni-coated) through a “strong” interfacial Ni-Ni bonding. As shown in Fig. 8, the fractions of such bonding ranged from 4.0% ($=2.8\% + 0.9\% + 0.3\%$) for the Under-weld, to 8.8% ($=4.8\% + 2.4\% + 1.6\%$) for the Normal-weld, and 14.5% ($=8.8\% + 4.8\% + 0.9\%$) for the Over-weld. It was so determined as “strong” Ni-Ni bonding for these fractions since the weld fractured on the pure Cu (of the innermost tabs) during the shear tests of the welds, knowing that the tensile strength for Ni (approximately 1 GPa) is much higher than that of pure Cu (ranging from 350 to 500 MPa). The rest of such fractions (i.e., 96.0% of the Under-weld, 91.2% of the Normal-weld, and 85.5% of the Over-weld) were considered “weak” interfacial Ni-Ni bonding.

(2) While a 100% “strong” Ni-Ni bonding seems to be desirable to achieve the best bonding strength, there exists two limiting factors. First, as discussed in (1), the upper limit of the bonding strength of a weld is determined by the Cu strength. Second, overly strong Ni-Ni bonding requires Over-weld conditions, which may cause over-thinning or circumferential fracture at weld perimeters as circled out at the left-upper corner in Fig. 7. In such Over-weld conditions and other extreme Over-weld conditions reported in Lee SS, et al. (2013), the weld strength might be lower or even significantly lowered than the Normal-weld conditions due to severe shear cut of the three Cu tabs by the horn at the weld pad perimeters. Therefore, an optimal welding conditions should be at a welding energy level equal to or lower than the “strong” Ni-Ni bonding energy.

(3) In manufacturing or even in R&D environment, weld tool misalignment often existed, such as shown in Fig. 8 with non-uniform “strong” Ni-Ni bonding areas on the three weld pads (i.e., Left, Middle and Right).

The lap shear tests were also performed for the cccC welds, with the test setup shown in Fig. 2. Only the bonding strength at the ccc|C interface is of concern (instead of at every interface) since it has been shown to be the weakest from both the experimental data and the heat dissipation analyses (Lee DK., et al. 2013). The shear load vs. the welding energy plot is shown in Fig. 9 at three different welding energy levels. The plot indicates that the peak shear force initially increases with the welding energy, and then starts to saturate at around 3200 J. Additional tests (Lee SS, et al. 2013) further discovered that the ccc|C lap shear strength actually decreases after certain energy levels (such as after 4000 J) due to the tab thinning and/or the weld spot circumferential fractures on the tabs. Note that the shear load is shown in a scaled form to protect proprietary data.

Figure 9. The shear load (at the ccc|C interface) vs. ultrasonic welding energy for cccC welds

4 Microstructural Analyses and Mechanical Tests of aaaC Welds

The photo images of representative aaaC welds from three different welding conditions are shown in Fig. 10. Note that the finer horn knurl pattern used was different from that of Cu welding.

Figure 10. The photo images of aaaC welds at three different energy levels (i.e., 1200 J for Over-weld, 600J for Normal-weld, and 300 J for Under-weld). The ultrasonic vibration is in vertical direction.

A representative cross-sectional view of as-welded samples is shown in Fig. 11, sectioned along the central line of long axis and polished without etching. It can be seen that:

- All three 0.2 mm Al layers were severely deformed, and the Al-Al interfaces were in wavy or curly. The amount of deformation decreased from outer to inner tabs. In particular, the Al inter-mixing is observed.
- At the aaa|C interface, the copper surface remained almost straight/flat. This is no surprising since Al is much softer and ductile, and thus more deformable than the 0.9 mm Cu.

Figure 11. Cross-sectional OM view of a normal-weld specimen sectioned along the long axis for the middle weld spot.

To view the Al tabs deformation and inter-mixing better, the polished sample in Fig. 11 was further electro-polished, as shown in Fig. 12 (top). The three tab layers and the interfaces were also schematically drawn in Fig. 12 (bottom). It is observed that:

- The grain size of the unwelded Al tab was about $50 - 67 \mu\text{m}$ since there were about 3 – 4 grains through the 0.2 mm tab thickness (shown in the very left portion of Fig. 11 (top)). After ultrasonic welding, severe plastic deformation and material flow completely destroyed the original grain structure. Recrystallization did not seem to occur because equiaxed new grains with well-defined grain boundaries were not observed ever in the most severely deformed region.
- The straight/flat boundaries between the three Al tab layers became highly wavy, curly, and discontinued after welding. Metal flow, inter-mix and interlock were seen.
- The aa|C interface, however, remained straight/flat. Therefore, the bonding mechanism between the innermost Al and the Cu layer was diffusion bonding.

Figure 12. (top) Sectioned and etched Al tabs interfaces, observed by OM (1.8% HBF₄ in distilled water; 20 vdc for 2 minutes; sample at anodic pole; Olympus OMG 3; polarizing light plus sensitive tint lens). Cu bus bar is not shown.

Fig. 13 further illustrates two distinct weld failure modes after an aaC weld was lap-shear tested. The first was the interfacial debonding between the innermost Al and Cu as evidenced by the large areas of the original rolling marks on the Cu sheet in Fig. 13 (a). This typically represents an under-welding condition due to insufficient welding energy. The second was the Al tab fracture, Fig. 13 (b) and (c), indicating adequate bonding between the sheets. Note that both normal-welding and over-welding conditions could produce adequate bonding; however, over-welding conditions typically create excessive circumferential cuts around the perimeters of the weld spots and therefore reduce the weld strength due to circumferential fracture (i.e., Mode III failure in fracture mechanics).

To further understand the bonding mechanism, SEM secondary electron images (Fig. 14(a)) and back-scattered electron images (Fig. 14(b)) were obtained and compared, along with EDX chemical analysis (Fig. 14(c) and (d)). The contrast of the secondary electron images was based on sample surface topology as well as atomic weight. The heavier metal reflected more electrons and displays lighter color with stereo surface topology feature, while the contrast of the back-scattered electron images was more sensitive to the atomic weight and gave a better black/white contrast. The two techniques yielded consistent results, and the chemical composition of the two regions was clearly confirmed by the EDX analysis.

Figure 13. (a) A fractured surface of aaC viewed on 0.9 mm Cu side; (b) The base metal shear fracture of aluminum tab; (c) The torn edge of the first (top) aluminum tab.

Figure 14. Micrographs of a local area of aa|C fractured surface. (a) secondary electron image; (b) back-scattered electron image, (c, d) EDX chemical analyses at location A of bark black area as Al and location B of grey area as Ni.

The fractured surfaces of all three weld spots and at all three energies are shown in Fig. 15 (top) as viewed on the 0.9 mm Cu mating surface. The Al side of the fractured three 0.2 mm Al tabs had a good match to the Cu side, but with severe tearing damage and shape distortion. The bonding areas between the Al tab and the Cu bus bar were traced at the bonding edges and filled with different colors, and the bonding area fractions over the full spot size were calculated with a user developed Matlab program (Fig. 15 bottom). Weld failure occurred as a combination of interfacial debonding between the innermost Al tab (anodized) and the Ni-coated Cu bus bar. The fracture occurred at the same spots for all three Al tabs as through-thickness tear (or Mode III fracture as defined in conventional fracture mechanics). For under-weld, the interface debonding was more predominant than the through-thickness tear. As the welding energy becomes higher, the interfacial bonding area becomes larger, resulting in higher weld strength until an upper limit that is equivalent to the maximum load carrying capability by through-thickness tear of the circumferences of the three weld spots. Similar to the Cu welding, the welding tool is also misaligned.

Figure 15. (top) OM observations of fracture of the Cu bonding surface aaa|C; (bottom) The remaining Al tabs from edge tearing for tab from top layer to the third layer in red, green and blue, respectively, and the remaining Al from interface shearing fracture in dark black spots (SEM).

The lap shear tests were also performed for the aaaC welds. Similar to the cccC welds, only the bonding strength at the ccc|C interface was tested. The shear load vs. the welding energy plot is shown in Fig. 16 at three different energy levels. The plot indicates that the shear load initially increases with the welding energy, then reaches a plateau and the peak at around 800 J, after which sees an decrease at over-weld conditions due to tab thinning and/or the weld spot circumferential fractures on the tabs. Note that in Fig. 16, the shear load is in a scaled form to protect proprietary data.

Figure 16. The shear load (at the aaa|C interface) vs. ultrasonic welding energy for aaaC welds

5 Conclusions

The purpose of this study is to better understand the mechanism of ultrasonic weld formation and failure to provide insight into welding process for quality improvement.

The formation mechanism of ultrasonic welds between three layers of battery tabs (either 0.2 mm thick Ni-coated Cu tabs or 0.2 mm thick anodized Al tabs) and the bus bar (Ni-coated 0.9 mm thick Cu) were studied, each at three different welding energy levels. Failure modes were examined and analyzed with optical, scanning electron microscope, and energy-dispersive X-ray spectroscopy.

It was concluded that while the weld formation mechanisms differ for Al and Cu welding, failure of the welds is always a combination of the interfacial debonding (between the innermost tab, either Cu or Al, and the Cu bus bar) and the through-thickness tear (i.e., Mode III failure as termed in fracture mechanics) of the tabs. More specifically, within the current welding process window,

For cccC welds (i.e., three layers of Cu tabs welded onto a Cu bus bar):

a) The primary welding mechanism was the Ni-Ni diffusion bonding instead of interlocking as evidenced by little metal inter-mixing.

b) Only a small fraction of the innermost Cu tab (Ni-coated) area was welded with the Cu bus bar (Ni-coated) through a “strong” interfacial Ni-Ni bonding, which is stronger than the Cu base metal, resulting in Cu base metal tearing apart at the fractured surfaces; while the rest of the area was through “weak” interfacial Ni-Ni bonding, with Ni residues over the fractured surface.

c) The weld strength is limited by the Cu strength, and an optimal welding condition is at a welding energy level equal to or lower than that required to produce a “strong” Ni-Ni bonding.

For aaaC welds (i.e., three layers of Al tabs welded onto a Cu bus bar):

a) The primary welding mechanism among the three Al tabs is the metal interlocking as evidenced by significant Al inter-mixing.

b) The highest possible weld strength is the maximum load carrying capability of the through-thickness tear of the circumferences of the three weld pads.

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