

Exploration of Interfacial Materials Chemistry Control to Improve Cu Wire Bonding Reliability

Kevin Antony Jesu Durai, Dinesh Kumar Kumaravel, John Alptekin, Logan Estridge, Shyam Nair,
Oliver Chyan*

Interfacial Electrochemistry and Materials Research Lab
University of North Texas, Denton, Texas
Denton, Texas, 76201, USA

Abstract

Copper wire bonding, with its advantages of higher electrical conductivity and better mechanical strength, has replaced gold wire bonding as a proven, cost-effective electrical interconnection solution for IC packaging for the past 15 years. Early Cu wire bonding development required overcoming several technical challenges, including bond pad damage caused by copper's hardness and brittleness relative to gold. A more chemistry-related challenge of using Cu as a bonding wire is its well-known reactivity with oxygen. An inert atmospheric envelope of forming gas surrounding bonding capillary was developed to prevent oxidation of Cu wire during electronic flame off to enable a strong bonding. Another more elusive materials-chemistry-related reliability challenge, with a typical low ppm occurrence, has been the chloride-induced corrosion defects between the Cu wire and Al bond pad. The opportunistic low-level chloride contaminations can originate from various points of the packaging manufacturing process flow, often rendering it un-trackable. In this paper, we present recent efforts to systematically control interfacial materials chemistry across Cu bonding wire, Cu/Al bimetallic contacts, and Cu_xAl_y intermetallic compounds with the aim to eliminate corrosion defects and improve the overall bonding reliability. The prevailing manufacturing solution is to utilize Pd-coated Cu bonding wire that can only partially mitigate the Cu_xAl_y intermetallic corrosion vulnerability. We utilized a real-time corrosion screening metrology to explore the underlying interfacial materials chemistry that drives vigorous corrosion between Cu wire and Al bond pad when exposed to a trace level of chloride contaminant. Combining with SEM, sensitive infrared spectroscopy, and electrochemical characterization, our data show that strategic surface modification on both Cu bonding wire and exposed Cu_xAl_y intermetallic can have a significant impact on reducing corrosion defect rates. The obtained mechanistic insights provide several new strategies enabled by a novel Cu-selective passivation coating technology to effectively mitigate Cu wire bonding corrosion defects. Implications for improving overall Cu wire bonding reliability will be presented based on these new approaches with low-cost and packaging-friendly advantages.

Key words

Cu wire bonding failure, Cu-Al intermetallics, Interfacial Materials Chemistry Control, Cu-selective passivation coating technology.

I. Introduction

Over the past decades, the use of Copper (Cu) as a wire bonding material has increased because of its superior electrical conductivity, 40% higher than gold, and other attractive mechanical properties. The main reason, however, was the cost reduction compared with gold wire bonding devices that were traditionally used. Several studies have explored the key technical challenges that Cu wire bonding faces as the replacement for gold wire bonding in integrated circuit (IC) packaging [1,2]. Harmon et al. pointed out the complicated requirements and one of the potential technical

obstacles involved in overcoming bond pad damage caused by Cu's hardness and brittleness compared to gold [2]. Jinzhi et al. reported that challenges like the reactivity of Cu towards oxygen have led to the development of forming gas ($\text{N}_2:\text{H}_2$) reductive atmosphere to prevent Cu oxidation while Cu wire bonding with Aluminum (Al) bond pad and also highlighted the advantages and challenges of Cu wire bonding technology [3].

In addition, the problem of halide-induced corrosion of Cu wires and Al bond pads has been found as one of the most significant reliability issues. Typically, these corrosion defects can also occur in the low ppm range from aggressive

low-chloride impurities in the packaging process [4,5]. One of the prevailing strategies to reduce these corrosion defects is the utilization of palladium (Pd) coated Cu wire (PCC), which only partially solves the corrosion problem [6-8]. Singh et al. reported earlier that various reliability testing done on PCC wire-bonded devices showed great intermetallic compound coverage with flat bonding interface but still resulted in reliability failure because of anomalous chloride levels [6]. Xu et al. studied the behavior of PCC wire and its impact on IMC growth in those wire bonding devices and reported the IMC corrosion problem [9]. Meanwhile, Lee et al. reported mechanical reliability studies comparing PCC and Cu wire, where both showed the same chloride induced corrosion resistance if Cu/Al IMC formation was properly formed [10]. Whereas Lim et al. and Wu et al. studied the corrosion performance of Cu/Al IMC with the effect of Pd addition, showing slightly improved corrosion resistance in Cu and Cu/Al samples [11,12]. However, controlling the distribution of Pd at the ball bond interface poses a challenge, leaving this metallurgical solution to inhibit the growth of intermetallic compounds (IMC) subject to potential imperfections due to the unpredictability of the ball melting process.

To better understand the mechanism of interfacial corrosion, numerous researchers have reported the behavior of Cu/Al IMC under different test conditions [13-16]. In particular, Kim et al. have explored the effect/importance of Cu/Al IMC formation on Cu wire and Al bond pad using mechanical test systems to study the failure mode of ball fractures and also reported the Cu_9Al_4 IMC as the predominant IMC at wire-bonding temperatures (150°C to 300°C) [14]. Meanwhile, Boettcher et al.'s further investigated the corrosion rate of the Cu/Al IMC species at the interface, finding that the Cu-rich phase corrodes faster compared to the Al-rich phase [17]. Liu et al. also studied the corrosion behavior of Cu/Al IMC in a chloride environment resulting in selective IMC corrosion with a high corrosion rate [4].

Further research has focused on evaluating the effects of epoxy molding compounds (EMCs) with two key factors, pH levels and Cl^- content, towards the substrate material properties on the reliability of Cu wire ball grid array (BGA) compounds [18] and developing new Cu wire materials coated with highly reliable and corrosion-resistant gold-palladium [19]. These studies offer valuable insights and potential directions for future research in our ongoing efforts to enhance the dependability and effectiveness of Cu wire bonding in microelectronics. Nonetheless, managing the interfacial materials chemistry to minimize the corrosion-related defect and enhance bonding reliability remains a complex challenge that the industry still faces.

In our prior research, we found that a Cu-Al bimetallic contact significantly accelerates galvanic corrosion. Using a

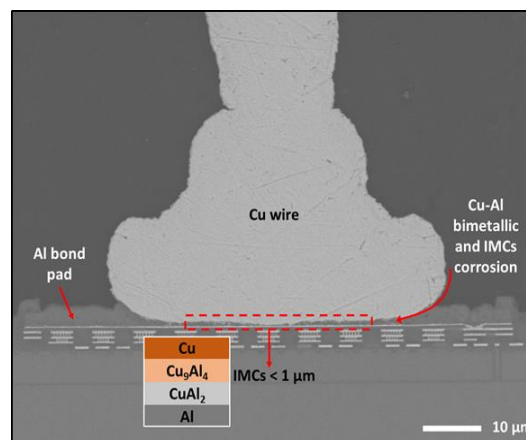


Fig. 1: Cross-section image of Cu-Al ball bond interface

micro-pattern test platform, we observed accelerated Al corrosion, even without IMCs present. Real-time corrosion screening showed corrosion progression on both the Cu bonding ball and Cu microdots on Al [20,21]. We identified hydrogen gas evolution as a critical factor for Cu ball bond lift-off and wire-bonded device failure. We also examined the role of IMCs corrosion in wirebond liftoff failure, a primary cause of device failure (Fig. 1). We demonstrated the effectiveness of a specific Cu-selective passivation coating and a corrosion inhibitor, “inhibitor A” in preventing hydrogen evolution reactions and bimetallic corrosion in the Cu/Al system. Our tests of coated devices showed full protection against Cu ball liftoff under extreme corrosion conditions [21]. These findings offer a solution to a significant reliability issue in wirebonding devices and guide future research on corrosion mechanisms and mitigation strategies.

In this paper, we address the chloride-based corrosion issue by discussing the development of a copper-selective passivation coating that offered over 90% protection according to our in-situ accelerated tests. We have also utilized our developed mechanistic understanding of the Cu-Al galvanic corrosion [21], and applied this understanding to investigate the corrosion behavior of the Cu-Al wire bond under an extremely corrosive environment of 600 ppm Cl^- aqueous solution. First, we performed a Zero Resistance Ammetry (ZRA) electrochemical study on the various galvanic couples possible in the Cu-Al wire bond when submerged in 600 ppm Cl^- over time in order to determine their current densities – both with and without addition of inhibitor A to solution. In addition, we constructed a Tafel plot of each species in the Cu-Al wire bond under 600 ppm Cl^- environment via Potentiodynamic polarization, to investigate the individual corrosion rate of each metal and IMC – with and without inhibitor A.

Alongside these electrochemical studies, Fourier transform infrared spectroscopy (FTIR) was used to monitor

the chemical bonding information of the metals and IMCs. In particular, the presence of chemical adsorption of the inhibitor (chemisorption), and also the development of Cu oxide growth with corrosion. To determine the effect of our passivation coating on real wire bonded devices, we employed our real-time corrosion screening metrology, where wire bonded devices (unencapsulated) were submerged in 600 ppm Cl^- solution for a total of 2 hours, and analyzed by an optical microscope for wire bond lift-off failure. Effective prevention of the corrosion mechanism requires a uniform and defect-free passivation coating application on the Cu wire, providing a physical barrier to hydrogen evolution. To test the uniformity of the developed inhibitor coating, we employed a micro-etching procedure on blanket Cu surfaces – both with and without passivation coating and investigated the surface modification via optical microscopy. Finally, we verified the adhesive strength between the passivation coating and the epoxy molding compound by tensile pull testing, as performed on blanket Cu surfaces – with and without passivation coating. FTIR spectroscopy was used to determine the breaking point in the Cu-inhibitor-epoxy interface.

II. Experimental

A. Electrode Preparation

Cu/Al IMC electrode preparation method was already reported in our previous work on studying Cu IMCs corrosion with a passivation strategy [23]. In our past work, pure Cu and Al alloyed with 0.5% Cu blanket wafers ($\sim 2 \mu\text{m}$) were used [23]. However, in this work, the Cu/Al IMC was prepared as follows: Si wafers were sputtered with a thin layer of Ti (for adhesion) and Ru (as a barrier layer between Cu/Al IMCs and Si). Then, using the Denton Vacuum Desktop pro sputtering unit, $\sim 400 \text{ nm}$ of Cu/Al IMCs were deposited by co-sputtering of Cu and Al. After sputtering, IMCs were annealed at 400°C in a nitrogen environment for 1 hour in a Linberg/Blue Mini-Mite Tube Furnace. Different compositions of Cu/Al IMCs were confirmed by a Rigaku SmartLab X-Ray Diffractometer by grazing angle XRD at a 0.5° incident angle relative to the surface plane.

B. Potentiodynamic Polarization and Zero Resistance Ammetry

All electrochemical experiments in this work were performed on a CH Instruments 760D Potentiostat. As discussed in our previous work [23], Potentiodynamic polarization (Tafel Plot) was done by a three-electrode system utilizing a Pt counter electrode and Ag/AgCl in a saturated KCl reference electrode with a scan rate of 1 mV/s . The potential window was set after 10 min of stable open circuit potential (OCP) when samples were immersed in the electrolyte solution. Zero Resistance Ammetry (ZRA) was used to measure the direct galvanic current of different

metals and IMCs in the bonding interface. In this setup, the more noble metal/IMC will act as the cathode, and the least noble metal/IMC will act as the anode. This ZRA measurement was carried out for 12 hours in various corrosive solutions. The current measured from this experiment will be converted to current density, calculated by dividing the obtained current by the surface area ($\sim 0.57 \text{ cm}^2$). ACS grade NaCl was used to prepare 600 ppm Cl^- solution with $18.2 \text{ M}\Omega\cdot\text{cm}$ ultrapure water from a Millipore Sigma water purification system.

C. ATR-IR of Post-Electrochemistry Electrodes

After electrochemistry, IMC electrodes were gently rinsed with ultrapure water and allowed to air dry in a clean room environment before taking ATR-IR (attenuated total reflectance – infrared) spectra. A Bruker Invenio FTIR (Fourier transform – infrared) spectrometer equipped with a DTGS (deuterated triglycine sulfate) detector and a Diamond ATR accessory was utilized for measurements. Background measurements were carried out on a set of as-prepared IMCs, Al, and Cu before measuring the respective post-electrochemistry electrodes. An integration of 300 scans was utilized for each background and sample measurement. Standard automatic atmospheric corrections, built into Bruker's OPUS software, were carried out to remove water vapor and carbon dioxide peaks resulting from imperfect background cancellations.

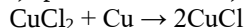
D. Corrosion Immersion Screening with and without Passivation Coating

For live microscopic corrosion screening and for optical images, a Nikon LV150N metallurgical microscope with a digital camera was used, as explained in our previous work [20]. For corrosion screening, 600 ppm Cl^- solution was used, as also utilized in the electrochemical experiment. Test devices with Cu-Al wire bonds (unencapsulated) from commercial sources were used for the entire corrosion screening studies, from that some of the devices will be coated by our Cu-selective passivation coating [22]. Scanning Electron Microscope (SEM) imaging was conducted in a JEOL IT200 at 15 kV under a high vacuum using a secondary electron detector (SED).

E. Micro-Etching Analysis to Determine Coating Uniformity on Cu Surface

To obtain proper corrosion protection on those Cu-Al wire bonding devices, the passivation coating should be uniform and defect-free. A micro wet-etching technique developed in our lab was used to provide information on localized coating defects and inhomogeneities. The chemistry of the chosen wet etchant is based on common printed circuit board (PCB) etching protocols in the semiconductor industry. The etchant

used is an acidic cupric chloride and hydrochloric acid solution. It works by oxidizing the Cu(0) surface (and in turn, reducing the Cu(II) species in solution).



The role of hydrochloric acid is to dissolve and solubilize the cuprous chloride formed on the surface so that the cupric chloride etchant can continue oxidizing the underlying Cu(0) [24]. For this work, a dilute concentration of 20 mM CuCl_2 solution was used as the Cu micro-etchant. Test Cu wafers with coatings were immersed for one minute in the micro-etchant solution and thoroughly rinsed with ultra-pure water immediately after removal.

F. EMC Adhesion Strength Testing

EME-G700 Type LG and EME-G631HA Type C Epoxy molding compounds (EMC) from Sumitomo Bakelite Ltd. were used for adhesion testing. Semiconductor-grade Copper lead frame substrates (Cu-LFs) were used to mimic the copper ball bond surface. Tensile pull-testing, evaluating the strength of the EMC adhesion, was conducted in the UNT Center for friction with the custom-made Mini tensile tester [30]. The Button shear test was conducted at the Sumitomo facility, and ATR-IR spectra were obtained using a Bruker Invenio FTIR spectrometer. For Adhesion testing, 8 x 8 mm square cutouts of Cu-LFs were taken and coated with inhibitor using our solvent-based process. Epoxy molding of Cu-LFs was carried out at 175°C under a heated pressure block and cured for another 2 hours at 175°C [31]. Then, those epoxy-molded (coated and non-coated) Cu-LFs went for pull testing.

III. Results and Discussion

A. Galvanic Corrosion Behavior in 600 ppm Cl^- and Influence of 0.1% wt/v Inhibitor

Galvanic corrosion current behavior for possible couples in the Cu-Al wire bond interface (Cu:Al, Cu: Cu_9Al_4 , Cu_9Al_4 :CuAl₂ and CuAl₂:Al) were studied using 600 ppm Cl^- solution with and without 0.1% wt/v inhibitor. The average current densities over 12 hours of immersion time are shown in Figure 2. The trend of the current densities for the control couples (without inhibitor) observed was as follows: Cu:Al > CuAl₂:Al > Cu_9Al_4 :CuAl₂ > Cu: Cu_9Al_4 . There is one key difference from our previous study in 100 ppm Cl^- solution [23], namely the Cu: Cu_9Al_4 couple now has only slightly less current density than the Cu_9Al_4 :CuAl₂ couple, instead of dramatically less. As the concentration of Cl^- in solution is greater in this study, the equilibrium of the anodic reaction is shifted towards the right, that is, towards more oxidation and a higher corrosion rate.

To explain the relative trends of current densities in Figure 2: the current density is highest in the Cu:Al couple, as these two metals have the highest difference in standard reduction

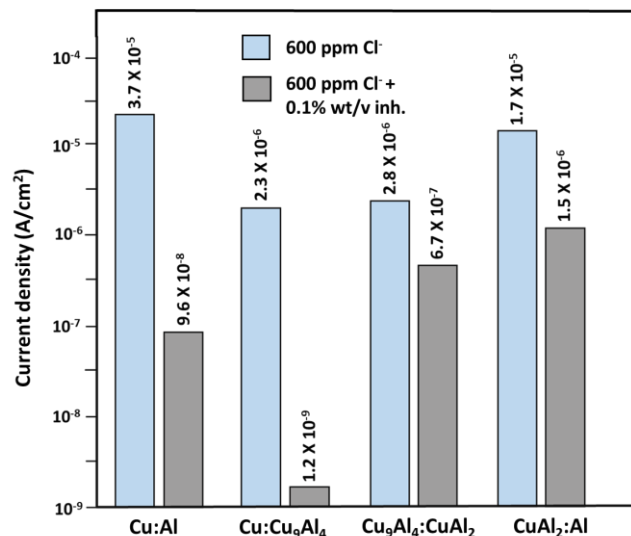


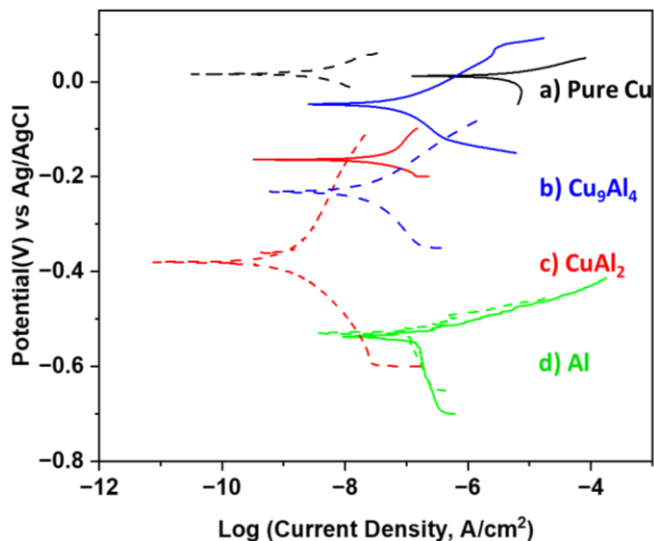
Fig. 2: Average current density values obtained from ZRA for 12 hours in 600 ppm Cl^- with and without inhibitor.

potentials (~ 2 V) of any two species present. The next highest is the CuAl₂:Al couple, which is higher than the other two IMCs due to its more Al-rich nature. As explained in our previous work [22], this higher current density in CuAl₂:Al is likely due to a dealloying process of CuAl₂ caused by localized pitting corrosion and selective micro-corrosion of the less noble Al within the CuAl₂, facilitated by chloride breaking down the protective Al oxides [25, 26].

As for the condition including 0.1% wt/v of inhibitor (Fig. 2), the trend of the current densities was: CuAl₂:Al > Cu_9Al_4 :CuAl₂ > Cu:Al >> Cu: Cu_9Al_4 . This trend was also different from our prior study with 100 ppm Cl^- solution [22], which was as follows: Cu:Al > CuAl₂:Al >> Cu_9Al_4 :CuAl₂ > Cu: Cu_9Al_4 . When comparing the differences between 100 ppm Cl^- and 600 ppm Cl^- it was revealed that the magnitudes of the current densities of Cu:Al, and Cu_9Al_4 :CuAl₂ are nearly identical, whereas the current densities of Cu_9Al_4 :CuAl₂ and CuAl₂:Al increased dramatically from 100 ppm to 600 ppm Cl^- . This further supports the hypothesis of the dealloying of CuAl₂ species, as facilitated by an extremely harsh Cl^- environment [26, 27].

Overall, the addition of 0.1% wt/v of inhibitor into 600 ppm Cl^- solution resulted in a reduction in current density for all the metal/IMCs, which shows the effectiveness of our inhibitor not only towards the Cu-Al galvanic system, but also with the different contact variation of metal/IMC's galvanic corrosion. Most importantly, the study demonstrated the corrosion protection potential of our developed passivation coating even in an extremely harsh 600 ppm chloride environment.

B. Individual Corrosion Behaviors in 600 ppm Cl^- and influence of 0.1% wt/v Inhibitor



Sample	E_{corr} (V)	I_{corr} (A/cm^2)
Cu in 600 ppm Cl^-	0.013	9.1×10^{-7}
Cu in 600 ppm Cl^- + 0.1% inh.	0.023	1.73×10^{-8}
Cu_9Al_4 in 600 ppm Cl^-	-0.06	1.3×10^{-8}
Cu_9Al_4 in 600 ppm Cl^- + 0.1% inh.	-0.23	3.5×10^{-9}
CuAl_2 in 600 ppm Cl^-	-0.16	1.1×10^{-7}
CuAl_2 in 600 ppm Cl^- + 0.1% inh.	-0.39	3.8×10^{-8}
Al in 600 ppm Cl^-	-0.51	4.8×10^{-6}
Al in 600 ppm Cl^- + 0.1% inh.	-0.53	3.4×10^{-7}

Fig. 3: Tafel plots of the metals and alloys utilized in this study in 600 ppm Cl^- (solid lines) and 600 ppm Cl^- + 0.1% inhibitor (dotted lines) for A) Pure Cu, B) Cu_9Al_4 , C) CuAl_2 , and D) Al.

Fig. 3 shows collective Tafel plots for individual metal in 600 ppm Cl^- solution with and without inhibitor. It is very evident that the addition of 0.1% wt/v of inhibitor has huge impact towards the decreased corrosion current density and individual corrosion rate for each metal and IMC. When the inhibitor was added, the corrosion potential (E_{corr}) of Cu shifted to a slightly more positive value, whereas for Al and Al-containing alloys E_{corr} shifted to more negative values. As mentioned in our previous work [23], inhibitor A suppresses bimetallic corrosion in the Cu:Al system by inhibiting the cathodic hydrogen evolution reaction on the Cu surface. This change in E_{corr} might be because of our passivation coating is only Cu selective, this leads the Al component in Cu/Al IMCs to undergo Al_xO_y formation to dominate the overall electrochemical behavior and move the E_{corr} to more negative side.

C. Interfacial Characterization of Electrodes After ZRA by ATR-IR Spectroscopy

After ZRA in chloride solution, ATR-IR of the electrodes (Fig. 4) demonstrated the growth of Cu(I) oxide, based on 650 cm^{-1} IR absorption peaks, on Cu and Cu/Al IMCs only and a relatively flat spectrum for Al. This is the expected result since the corrosion solution doesn't contain any organic species able to bind strongly to the surface of the electrodes. Without any corrosion inhibitor in solution, the surfaces of the electrodes readily corrode and oxidize, as both Cu and Al are well-known to undergo pitting corrosion in the presence of chloride [28,29]. The evolution of Cu(I) oxide (650 cm^{-1}) could have originated from the solution-assisted oxidation of Cu due to dissolved oxygen and adsorption of $-\text{OH}$ on Cu in water [28]. Additionally, CuCl surface products are hypothesized to react with water to form Cu(I) oxides at low chloride concentrations [29].

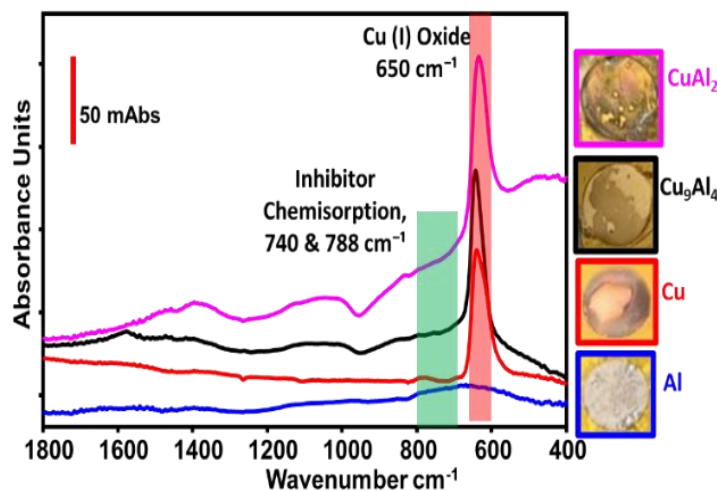


Fig. 4: ATR-IR spectra of the metals and alloys after ZRA in Cl^- solution. From top to bottom: CuAl_2 , Cu_9Al_4 , Cu, Al.

After ZRA measurement in 600 ppm chloride with the addition of 0.1% inhibitor, ATR-IR collected on the electrodes (Fig. 5) demonstrated almost no Cu oxidation, making the electrodes appear less corroded. Furthermore, it was shown that the inhibitor prevented corrosion on the Cu and Cu-Al IMCs, and verified to chemisorb onto the IMCs surfaces via the presence of the peaks located at 740 cm^{-1} and 788 cm^{-1} wavenumbers, even despite thorough rinsing of electrodes before ATR-IR measurement. The inhibitor also suppressed the formation of Cu(I) oxides compared to the electrodes submerged in only chloride. Thus, the inhibitor

prevents both the anodic dissolution of the electrodes and their Cu oxide formation. Another important observation, the bottom ATR-IR spectrum of Al electrode surface showed no sign of adsorbed inhibitor. The absence of adsorbed inhibitor on Al metal surface suggests the inhibitor adsorption is Cu-selective.

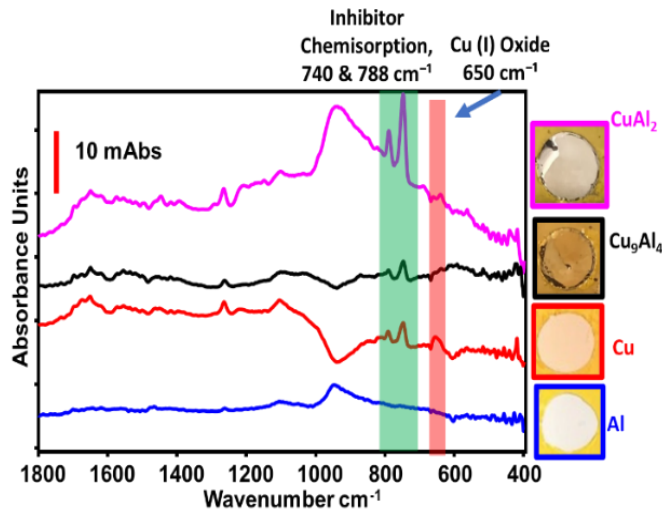


Fig. 5: ATR-IR spectra of the metals and alloys after ZRA in $\text{Cl}^- + 0.1\%$ inhibitor. From top to bottom: CuAl_2 , Cu_9Al_4 , Cu, Al.

D. Corrosion Immersion Screening on Cu-Al Wire Bonded Devices

To evaluate the corrosion protection achieved by our passivation coating, two distinct samples were compared: the as received (non-coated) Cu-Al device and the device with passivation coating. The corrosion immersion screening was conducted by placing the two distinct sample devices in a severe corrosive environment of 600 ppm Cl^- aqueous

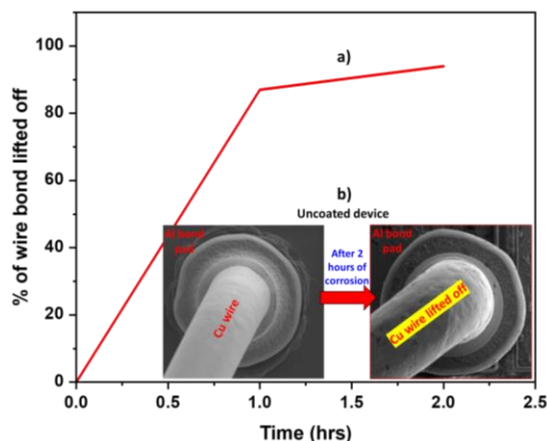


Fig. 6: Corrosion immersion screening of non-coated Cu-Al wire-bonded device in 600 ppm Cl^- . a) Plot shows % of wire-bonds lifted off with respect to immersion time in corrosive medium and b) SEM image of wire bond before and after 2 hrs of corrosion screening.

solution, which is much more corrosive than the typical operating condition employed in thermal stress testing by the IC packaging sector. After 2 hours of corrosion screening, non-coated devices showed over 93% of severe Cu wire bonding lift-off (Fig. 6(a)). As discovered in our previous work [21], galvanic corrosion was initiated between the Cu/Al bonding interface with Al_2O_3 dendrite formation and H_2 gas evolution. Fig. 6(b) demonstrates a Cu wire lift-off after Al bond pad corrosion. Subsequently, the corrosion expands towards the Cu/Al IMCs and causes a weak Cu/Al bond interface, culminating in Cu wire lift-off and shifts away from Al bond pad. On the other hand, Cu-Al devices with passivation coating encountered less than 15% of wire lift-off (Fig. 7(a)) after 2 hours exposure to highly corrosive 600 ppm Cl^- aqueous solution. In our previous work [23], we successfully achieved 0% wire bond lift-off after 2 hrs

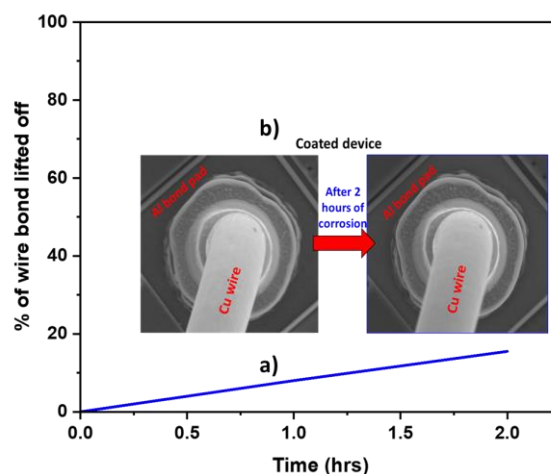


Fig. 7: Corrosion immersion screening of coated Cu-Al wire-bonded device in 600 ppm Cl^- . a) Plot shows % of wire-bonds lifted off after 2 hrs immersion in corrosive medium and b) SEM image of wire bond before and after 2 hrs of corrosion screening.

submersion in 100 ppm Cl^- solution. SEM images from Fig. 7(b) display a coated device without any corrosion on the Al bond pad and no detachment of the copper wire. By preventing the crucial process of cathodic H_2 evolution through the application of our passivation coating, the galvanic-induced corrosion caused between the Cu/Al bonding interface was significantly retarded [21].

E. Evaluation of Coating Uniformity on Cu Surface Using Micro-Etching Analysis

Using this micro-etching screening method, it is possible to detect whether there are any defects or non-uniformities present on the coated Cu substrate. Two different samples were chosen for this analysis, one bare Cu substrate (used as a control) and the other a passivation layer coated Cu

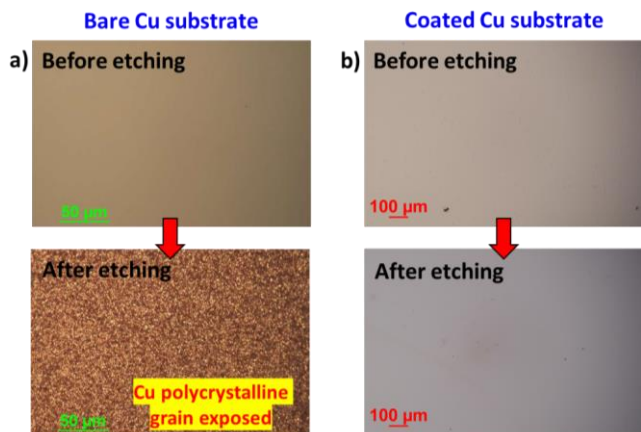


Fig. 8: Optical image before and after micro-etching a) Bare Cu and b) Coated Cu.

substrate. Figure 8(a) provides the optical images of the bare Cu substrate, quickly revealing its exposed polycrystalline grain structure after 1 min of etching. Whereas the Cu-specific micro-etchant does not affect the inhibitor-coated Cu (Fig. 8(b)), which confirmed that the optimized coating on Cu was free from defects and had uniform coverage over the Cu surface.

F. EMC Adhesion Strength Testing with Passivation Coating.

Test no.	Test method	Surface finishing	Pull strength (N)
1	Tensile pull test	Bare Cu (controlled)	30
2		Passivated Cu	16
3	Button shear test	Bare Cu (controlled)	134.6
4		Passivated Cu	61.1

Table 2: Adhesion pull strength with EMC on bare Cu-LFs and passivation layer coated Cu-LFs.

From the tensile pull strength testing, we observed that the molded bare Cu demonstrated a reliable pull strength of over 30 N, while the coated gave only around 16 N, which is considerably lower. Table 1 shows a similar trend obtained in the Button shear test. Figure 9 shows that post-pull test ATR-IR spectroscopy was obtained with several distinct absorption bands, including aromatic C-H stretching and Si-CH₃ within the 700-800 cm⁻¹ region and a significant broad absorbance between 900-1150 cm⁻¹. This latter region reveals peaks at 1020 cm⁻¹ and a shoulder at 1100 cm⁻¹, corresponding to Si-O-Si and Si-O-C bonding, respectively, consistent with known silica fillers in EMC. The differential spectra between the coated and non-coated samples showed the presence of passivation coating on the epoxy, which was confirmed by the 788 cm⁻¹ and 740 cm⁻¹ peaks. Thus, proving the breakage point is in between the coating layer, not in the

interface between EMC and the coating layer. This further demonstrates the adhesion capability of our coating with EMC.

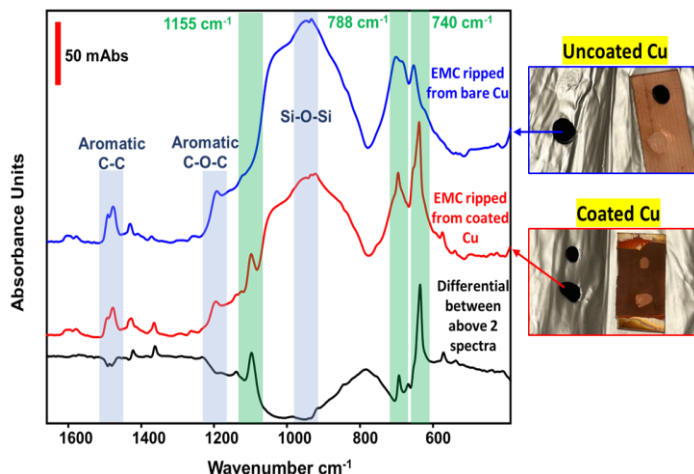


Fig. 9: ATR-IR spectra showing coating leftover on the EMC ripped from the sample surface.

III. Conclusion

In this work, the challenges and solutions of improving Cu wire bonding reliability against chloride-induced corrosion defects were studied through interfacial materials chemistry control.

- The galvanic corrosion behavior of various couples of Cu/Al and IMCs were investigated in a highly corrosive 600 ppm Cl⁻ environment. In general, adding a mere 0.1% inhibitor to the solution reduced the average current density values obtained from the ZRA technique after 12 hours of immersion time by several orders of magnitude over the control. Potentiodynamic polarization also supports the ZRA results where the I_{corr} reduces with the addition of the inhibitor.
- Post FTIR analysis on Cu/Al IMCs samples that underwent ZRA in 600 ppm Cl⁻ solution with added inhibitor shows a lack of Cu(I) oxide and chemisorption of inhibitor on Cu/Al IMCs.
- Corrosion immersion screening in 600 ppm Cl⁻ solution supported our previous work results where only 15% of Cu wire lift-off was recorded.
- Micro-etching analysis on inhibitor coating blanket Cu surface confirmed the coating's uniformity and defect-free nature.
- We have also verified the adhesive strength between the inhibitor and the EMC through various pull testing and used FTIR spectroscopy to determine that the breaking point in the Cu-inhibitor-EMC interface is not between EMC-inhibitor, however, the inhibitor's chemical bonding signature was found on both the EMC and Cu surface.

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