



Formation of Interfacial Reaction Layers in Al₂O₃/SS 430 Brazed Joints Using Cu-7Al-3.5Zr Alloys

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Abstract: The formation of interfacial reaction layers was investigated in an α -Al₂O₃/430 stainless steel (SS430) joint brazed using a Cu-7Al-3.5Zr active brazing alloy. Brazing was conducted at above its eutectic temperature of 945 °C and below liquidus 1045 °C, where liquid and solid phases of the brazing alloys coexists. At 1000 °C, the liquid phase of the brazing alloy was wet onto the α -Al₂O₃ surface. Zr in the liquid phase reduced α -Al₂O₃ to form a continuous ZrO₂ layer. As the dwell time increased, Zr in the liquid phases near α -Al₂O₃ interface was used up to thicken the reaction layers. The growth kinetics of the layer obeys the parabolic rate law with a rate constant of 9.25 × 10⁻⁶ cm·s^{-1/2}. It was observed that a number of low yield strength Cu-rich particles were dispersed over the reaction layer, which can release the residual stress of the joint resulting in reduction of crack occurrence.

Keywords: ceramic-metal joining; brazing; active brazing alloy; interface; zirconia; microstructure

1. Introduction

Heterogeneous metal-ceramic joints have been found in a number of applications such as electronic devices, automobiles, and high-temperature batteries [1–7]. One of popular methods for metal to ceramic boning is brazing, in which an active brazing alloy (ABA) is placed between the ceramic and metal components, followed by heating for a certain period of time to induce a chemical bonding at the interface [6,8]. Active brazing alloys typically consist of three main alloying elements, which are called an *ABC* alloying system. *A* is the parent constituent element that determines the baseline of the brazing temperature, *B* is added to adjust (typically lower) the liquidus temperature of the alloy to below the melting point of the substrates, and *C* is applied to help to introduce a highly reactive metal such as Ti, Zr, V, or Cr [2,9–13].

One of the most widely used active brazing alloys is the Ag-Cu-Ti alloy, which consists of an Ag-Cu eutectic system containing 1.5–5.4 wt.% Ti as an active element. Because a brazed joint using this type of alloy provides a high bonding strength and good oxidation resistance at high temperatures of up to 600 °C [14,15], the alloy has been applied to a wide variety of energy and automotive applications under high-temperature environments. However, in the case of a liquid Na environment such as a sodium (Na) beta-alumina battery (NBB), specimens using an Ag-Cu-Ti alloy have shown to fail within less than 100 h because the Ag in a Ag-Cu-Ti alloy is continuously dissolved into liquid Na at 350 °C, leaving a porous Cu-Ti structure [16].

In developing a stable high-temperature brazing alloy resistant to liquid Na, previous studies have suggested adopting Cu-7Al-xTi and Cu-7Al-xZr (x = 2.5, 3.5 and 4.5) alloys, which consist of



a non-dissolving element into Na (Cu, Al, Ti and Zr) and have high melting points of 965–1033 °C and 933–1051 °C, respectively. When the α -Al₂O₃/SS430 specimens were brazed using the alloys at 1000 °C, where the liquid and solid phases of brazing alloy coexisted, crack-free joints were observed for joints using only Cu-7Al-*x*Zr alloys. In addition, it was proposed that the minimum residual stress causing cracking in the joint can be expected using the hardness of the brazed seam [17].

For the metal–ceramic joint using the brazing alloy such as Ag-Cu-Zr, AgZr, and CuZr systems, formation ZrO₂ (zirconia) oxide is known to be the main factor in providing good adherence [17–23]. The majority of studies have been conducted above the liquidus temperature to improve the wettability of the brazing alloy and chemical reaction. However, in our previous experiment [17], it is interesting that although the brazing alloy was not fully liquefied, a tight bonding was obtained without cracking at the joint. In this study, to better understand the bonding mechanism of liquid–solid state bonding for the Cu-7Al-3.5Zr alloy, formation of the interfacial reaction layer was investigated in terms of microstructure analysis as the bonding temperature and dwell time increased.

2. Materials and Methods

A 430 stainless steel (SS430) plate (12 mm × 12 mm × 3 mm, Fe-14Cr-0.12C-1Si-1Mn in wt.%, Dongwon Special Steel Co., Busan, Korea) and circular α -Al₂O₃ (97% purity, 10 mm in diameter × 5 mm in length, Ujuoo Co., Busan, Korea) were machined for the substrates. The substrates were ground with 600 grit (P1200) SiC paper and cleaned in ethanol using an ultrasonic machine for 10 min. The Cu-based brazing alloy, Cu-7Al-3.5Zr was cut into 10 mm × 0.4 mm sections using electronic wire cutting, ground to a thickness of 0.2 mm using the same abrasive paper, and then ultrasonically cleaned for 5 min. The eutectic (T_E) and liquidus (T_L) temperatures of the alloys were determined using a differential scanning calorimeter (DSC, Universal V4.5A, TA Instruments, Newcastle, DE, USA) at 945 and 1045 °C, respectively.

A thin brazing filler of 200 μ m was placed between the SS430 and α -Al₂O₃, as shown in Figure 1. A pressure of 2.00 g·mm⁻² was loaded on top of the α -Al₂O₃ perpendicularly to the bonding plane in order to fix the brazing alloy. The specimens were placed in a vacuum furnace at a high pressure of 1.0×10^{-5} Torr and heated to 950 °C, 970 °C, and 1000 °C at a heating rate of 10 °C·min⁻¹. The dwelling times were 0, 10, and 30 min at 1000 °C and cooled at a rate of 2 °C ·min⁻¹. Here, "0 min" indicates that no holding time at the brazing temperature was applied. Under each temperature and time couple, three duplicated specimens were prepared. Cross-sectional micrographs of the brazed joint were etched into a solution made up of 1 mL HCl + 1 g FeCl₃ + 200 mL distilled water. The microstructures were characterized using optical microscopy (OM, BX-51M, Olympus, Tokyo, Japan), secondary electron microscopy (SEM, SUPRA45, Carl Zeiss, Oberkochen, Germany) with an energy diffraction X-ray (EDX), a field-emission electron probe micro-analyzer (FE-EPMA; JEOL (Tokyo, Japan) JXA-8530F) and X-ray diffraction (XRD; Ultima IV, Rigaku, Tokyo, Japan). For the interface analysis of the reaction layer, a small area of the specimen, 10 µm thick, was milled using a focus ion beam (FIB; FEI (Hillsboro, OR, USA) Scios focused ion beam) and the diffraction pattern and chemical composition for the phases were analyzed using a field-emission transmission electron microscopy (FE-TEM; TALOS (Hillsboro, OR, USA) F200X).



Figure 1. Schematic illustration of the brazed specimen.

3. Results and Discussion

3.1. Reaction of Cu-7Al-3.5Zr Alloy with Substrates

Figure 2 shows the brazing cycle and the corresponding cross-sectional optical micrographs (OM) of α -Al₂O₃ / SS430 joints brazed using Cu-7Al-3.5Zr alloy. When the holding temperature was below 1000 °C, the α -Al₂O₃ substrate was not joined with the brazing alloy. However, for the specimens brazed at 1000 °C between the eutectic (*T_E*) and liquidus (*T_L*) temperatures, α -Al₂O₃ and the brazing alloy were joined without defects such as discontinuities, pores, and cracks, indicating that a sufficient chemical interaction occurred at the interface during the brazing. Figure 3 shows the corresponding elemental mapping of Cu, Al, Zr, and Fe using FE-EPMA for the specimen brazed at 1000 °C for 10 min. It can be seen that in Figure 3c, the segregation of Zr is clear at both interfaces, but Zr intensity at the α -Al₂O₃-Cu-7Al-3.5Zr interface is extremely high. Fe was not detected at all in the brazed seam, indicating that Fe was not dissolved into the Cu because the solubility of Fe into Cu is only 1.5 wt.% [23].



Figure 2. (a) Brazing heating cycle and the corresponding optical micrograph (OM) of a α -Al₂O₃/Cu-7Al-3.5Zr/SS 430 joint brazed under (b) as-cast conditions and at (c) 950 °C for 0 min, (d) 970 °C for 0 min, and (e–g) 1000 °C for 0, 10, and 30 min, respectively.



Figure 3. FE-EPMA mapping of Cu, Al, Zr, and Fe elements for the α -Al₂O₃/Cu-7Al-3.5Zr/SS 430 joint brazed at 1000 °C for 0 min (**a**–**d**).

Enlarged SEM images for the two dotted areas (X and Y in Figure 2) are shown in Figure 4a,b, respectively, and six distinct phases were characterized using spot quantitative analyses (in at.%) through an EDX analysis, as represented in Figure 4c. It is observed that the microstructure of the as-cast Cu-7Al-3.5Zr alloy consists of a primary α -Cu solid solution (SS, #3) containing 12-13 at.% Al and eutectic phases (#1 and #2) along the entire α -Cu SS boundary. Zr was not detected in the α -Cu SS because the solubility of Zr in Cu at room temperature is negligibly small, whereas that of Al reached approximately 13 at.% [17]. It can therefore be inferred that Zr is used to form intermetallic compounds and Al is soluble in α -Cu. The eutectic phases are an eutectic structure containing α -Cu (#2) and an intermetallic AlCu₂Zr compound (#1), which is identified in Figure 5a through an XRD diffraction pattern. The crystal structure of AlCu₂Zr is a face-centered-cubic (FCC). The microstructure of the brazed seam bonded at 1000 °C for 0 min consists of the primary α -Cu SS (#4) and eutectic structure alternating between the α -Cu (#5) and AlCu₂Zr (#6) phases, which were also confirmed using the XRD analyses in Figure 5a. It is interesting that the same phases were detected for the as-cast and

brazed seams at 1000 °C for 0 and 30 min. However, the area fraction of the eutectic phases tends to decrease with an increase in temperature and dwell time during brazing, as shown in Figure 5b. Here, the area fraction of eutectic phases was measured using a pixel counting method from the OM images in Figure 2b–g. The values were averaged for the three duplicated specimens brazed under the same conditions and represented with error bars. The results indicated that the eutectic phases were used to react with both substrates.



Figure 4. Magnified secondary electron (SE) images of (**a**) as-cast brazing alloy and (**b**) seam brazed at 1000 °C for 0 min, and (**c**) energy-dispersive spectrometer (EDS) spot quantitative analysis (at.%) of the six distinctive phases shown in Figure 3a,b.



Figure 5. (a) X-ray diffraction patterns of as-cast brazing alloy and brazed seam at 1000 °C for 0 and 30 min, respectively and (b) the area fraction of the AlCu₂Zr phases in brazed seam at the different bonding conditions shown in Figure 2.

As observed in the elemental distribution within the FE-EPMA results in Figure 3, the Zr element was distinctively segregated within the vicinity of the α -Al₂O₃/Cu-7Al-3.5Zr interface, indicating that Zr in the AlCu₂Zr was used to react with α -Al₂O₃. It is known that the redox reaction between Zr and α -Al₂O₃ is thermodynamically favorable at above 675 °C, which corresponds roughly to the melting point of Al (660 °C) [24]. Therefore, it is expected that ZrO₂ oxide can be favorably developed at 1000 °C ($\Delta G = -4.5 \text{ kcal} \cdot \text{mol}^{-1}$), although the brazing alloy was not fully liquefied to be wet on the α -Al₂O₃. A detailed microstructure characterization of the reaction layer and its formation process will be further examined in Section 3.2.

3.2. Formation of the Interfacial Reaction Layers Brazed at 1000 °C

To understand the continuous reaction between Zr and α -Al₂O₃ during brazing, the interfacial microstructure was investigated as the bonding process proceeds. Images of secondary electron (SE) micrographs of the reaction layers at the α -Al₂O₃ / Cu-7Al-3.5Zr interface of the joint are shown in Figure 6, with a magnified view provided below. As the bonding temperature and dwell time increased, (1) continuous reaction layers between α -Al₂O₃ and Cu-7Al-3.5Zr could be clearly seen above 1000 °C, and (2) small dotted particles appeared below the reaction layer, the volume fraction of which gradually reduced.





Figure 6. Secondary electron (SE) images of the reaction layers at the α -Al₂O₃ / Cu-7Al-3.5Zr interface of the joint brazed at (**a**) 980 °C for 0 min, (**b**) 1000 °C for 0 min, (**c**) 1000 °C for 10 min, and (**d**) 1000 °C for 30 min. Magnified views of the dotted rectangles and their schematics for the microstructural characteristics are suggested below.

To identify the initial reaction at α -Al₂O₃/ SS430 interface, a magnified image of the interfacial microstructure brazed at 980 °C for 0 min was investigated. Figure 7 shows a high-angle annular dark-field (HAADF) image of area X in Figure 6a along with the corresponding elemental mapping for Cu, Al, Zr, and O. Five distinctive phases are labeled as #1–#5 in Figure 7. It should be noted that the reaction layer at this interface consists of a Zr- and O-rich continuous layer (#3) and Cu-rich island-morphological phases (#2) inside the layer. To estimate the composition and identify the phase, spot-quantitative analyses using EDX were carried out, the results of which are given in Table 1. In addition, their corresponding diffraction patterns (#1–#5) are shown in Figure 8. It was determined that a continuous reaction layer composed of monoclinic ZrO₂ (a = 0.5190 nm, b = 0.5213 nm, and c = 0.5310 nm) of 2 µm thickness and a small amount of Al-containing Cu particles of approximately 0.1–0.4 µm in thickness developed. The AlCu₂Zr intermetallic compound (#4, face-centered crystal, FCC) and α -Cu (#5, face-centered crystal, FCC) were present immediately below the ZrO₂ reaction layer, which indicates that the liquefied eutectic reacted with the α -Al₂O₃ and solidified after the brazing was finished.



Figure 7. (a) High-angle annular dark-field (HAADF) image of area X in Figure 6, and (**b**–**e**) elemental mapping for Cu, Al, Zr, and O, respectively.

No.	Cu	Al	Zr	0	Possible Phase
#1	-	41.5	-	68.5	α -Al ₂ O ₃
#2	89.1	10.9	-	-	α-Cu
#3	-	-	42.3	57.7	ZrO_2
#4	57.9	22.1	20.0	-	AlCu ₂ Zr
#5	89.5	10.5	-	-	α-Cu

Table 1. EDX analysis of the distinctive five phases (#1–5) shown in Figure 7 (at.%).

(a)	·	(b) . (ī11ī)	(c)	(d) (202) =	(e)	• •
•	(012)	. (111)	(111) (1ī0)	(022)		(020)
	(014) Al ₂ O ₃ [100]	(200) Cu-FCC [011]	ZrO ₂ -Monoclinic [112]	AICu₂Zr-FCC [Ī11]	Cu	(200) -FCC [001]

Figure 8. (**a**–**e**) Corresponding diffraction patterns obtained from areas labelled #1–#5 in Figure 7, respectively.

It is interesting to note that as the bonding temperature and dwelling time proceeded, the volume of the AlCu₂Zr intermetallic compounds near the interface gradually decreased and nearly diminished at 1000 °C after 30 min. In contrast, the thickness of the ZrO₂ reaction layer increased with an increase in time. This microstructural behavior is schematically described in Figure 6. To quantify the growth kinetics of the ZrO₂ layer, the thicknesses of three random areas were averaged for each dwell time. Their averaged values along with the standard deviations (μ m) are plotted in Figure 9 as a function of the squared time (in seconds). This illustrates that the growth rate of the ZrO₂ layer follows the parabolic rate law with an *R*² value of 0.982, where the rate constant at 1000 °C was determined to be 9.25 × 10⁻⁶ cm·s^{-1/2}. This parabolic relationship infers that the growth of the layer was followed by diffusion. Previous studies have suggested that the growth of ZrO₂ was controlled by the diffusion of oxygen through the ZrO₂ layer [25]. This phenomenon can be supported in that, as shown in Figure 6, the reaction front between Zr and α -Al₂O₃ has moved to inside of the α -Al₂O₃ → 2Al + 3O.



Figure 9. Thickness of ZrO_2 reaction layer (in μm) as a function of the dwelling time (s^{1/2}).

An X-ray diffraction (XRD) pattern for an inter-diffusion layer at the interface between α -Al₂O₃ and Cu-7Al-3.5Zr is shown in Figure 10. To prepare the specimens for the XRD analyses, as-brazed specimens were ground down from the SS430 side parallel to the bonded plane until the ground plane reached the α -Al₂O₃ / Cu-7Al-3.5Zr interface. ZrO₂ and Cu were identified at the interface corresponding with TEM analysis in Figure 8. It is observed that Cu peak corresponded to the Cu particles (Figure 8b) in the ZrO₂ layer and α -Cu solid solution (Figure 8e) in the brazing alloy. The Al₂O₃ peak was affected by the base substrate, which was determined in Figure 8a. The formation of the ZrO₂ phase can be explained as follows: (1) Al₂O₃ was decomposed into Al and O, namely,

Al₂O₃ (s) \rightarrow 2Al (s) + 3O (g), and (2) Zr in the liquefied eutectic reacted with decomposed α -Al₂O₃ forming ZrO₂, namely, Zr (s) + O₂ (g) = ZrO₂ (s). The redox reaction of Zr with Al₂O₃ (3Zr + 2Al₂O₃ = 3ZrO₂ + 4Al) is known to be thermodynamically favored because the Gibbs energy for the reaction is negative at 1400 K ((ΔG = -49.9 kJ, at 1400 K) [26]. When considering the same reaction process at a different bonding temperature, namely, 1000 °C (1273 K) in the present study, the Gibbs free energy for the reaction at 1000 °C (1273K) was -21.2 kJ, indicating that the reaction was also thermodynamically favorable. The Al released through the redox reaction diffused into the solid Cu solution, as can be expected given the fact that 11–13 at.% of Al was dissolved into Cu. Also, it is considered that the existence of small Cu particles in the reaction layer was developed by the diffusion of Cu in the brazing alloy into the already formed ZrO₂ layer [23].



Figure 10. X-ray diffraction (XRD) pattern for an inter-diffusion layer at the interface between α -Al₂O₃ and Cu-7Al-3.5Zr showing that the presence of ZrO₂ is evident at the interface.

It should be noted that the reaction layer consists of ZrO_2 oxide and α -Cu alternating between these two phases. In contrast, the interfacial microstructure using a Ti-containing brazing filler metal has continuous and flat Ti₂O or Ti₃Cu₃O reaction layers at the Al₂O₃ interface [14,27]. This difference is considered to be because that Ti-containing alloys has better wettability with the lower contact angles on the Al₂O₃ rather than Zr-containing one [18]. Also, the alternative structure of $ZrO_2 + Cu$ layer can be properly explained by the differences in density between Cu and ZrO_2 . The densities of ZrO_2 and Cu were 5.7 and 8.96 g·cm⁻³, respectively [20], which could account for the buoyant forces at the reaction front. It is considered that ZrO_2 adjacent to the α -Al₂O₃ surface was transported away from the reaction front due to the differences in density of Cu and ZrO_2 and separately aggregated behind.

For the ceramic-metal joint, the residual stress caused by mismatches in the thermal coefficient expansion (CTE) between the ceramic and metal components should be controlled in order to achieve good mechanical properties [6,8,28–30]. However, when selecting an intermediate layer between the ceramic and metal, not only the differences in the thermal expansion mismatch but also the mechanical characteristics, such as the yield strength, should be considered [28,29]. Figure 11 shows a magnified HAADF image of area X in Figure 7 for a joint brazed at 980 °C for 0 min. It was observed that in the ZrO₂ reaction layer, a number of Cu particles were widely distributed. Based on the microstructural characteristics of the layer, it is expected to decrease the residual stress considering the CTE and yield strength (σ_y), which for α -Al₂O₃, Cu-7Al-3.5Zr, SS 430, ZrO₂, and α -Cu are listed in Table 2. For the thermal and mechanical properties of the Cu-7Al-3.5Zr brazing alloy and α -Cu, a compositionally similar Cu-7Al-2Fe alloy (C61,400) and Cu-8Al (C61,400) were referred to [17,31,32]. ZrO₂ has an intermediate CTE value between the α -Al₂O₃ and brazing alloy (Cu-7Al-3.5Zr), and Cu particles have a low yield strength, which can have an effect on the relaxation of residual stress between α -Al₂O₃ and the brazing alloy providing a sound brazed joint without cracking.



Figure 11. Magnified high-angle annular dark-field (HAADF) image of deflected area X in Figure 7 for the joint brazed at 980 °C for 0 min.

Materials	α (10 ⁻⁶ K ⁻¹) CTE	σ_y (MPa) Yield Str.
α -Al ₂ O ₃	7.0	-
SS 430	10.4	310
Cu-7Al-3.5Zr	16.2	133
ZrO_2	10.0	-
α-Cu	18.0	185

Table 2. Thermal and mechanical properties of the materials used in the simulations (at $25 \degree C$).

Figure 12 shows an SEM image of the interface between Cu-7Al-3.5Zr and SS430 in the joint brazed at 1000 °C for 10 min. To identify the interfacial microstructure, a thin section was created using focused ion beam (FIB) milling and was then investigated through transmission electron microscopy (TEM) techniques. Figure 12b shows an HAADF image of the dotted area X in Figure 12a and the corresponding elemental spectral analysis for Cu, Al, Fe, Zr, and Cr is shown in Figure 12c. It can be clearly observed that a Zr-rich reaction layer of about 0.2 μ m thickness, (Fe,Cr)_x(Zr)_y, is continuously present below the brazing alloy (AlCu₂Zr and α-Cu). An Al-rich sub-layer of about 0.1 μ m thickness, (Fe,Cr)_x(Al)_y, was discontinuously formed beneath the (Fe,Cr)_x(Zr)_y layer. This layer was also identified through a similar study involving the identification of the interface between Invar (Fe-Cr-Co) alloy and Ag-Cu-Zr brazing alloy [33]. It should be noted that Zr in the brazing alloy was used to form a continuous Zr rich $(Fe,Cr)_x(Zr)_y$ layer. In addition, Al in the brazing alloy and/or excess Al from the redox reaction, was used to border the discontinuous Al-rich sub-layer, $(Fe,Cr)_x(Al)_y$. It is expected that the two layers enhanced the adherence of the brazing alloy and SS430. Unreacted Zr in the bulk remained as AlCu₂Zr in the brazing seam. However, when comparing the thicknesses of the two reaction layers (3–6 μ m for ZrO₂ and 0.5 μ m for (Fe,Cr)_x(Zr)_y), Zr was mainly used to form the ZrO₂ reaction layer, which parabolically grew with an increase in the holding time, whereas $(Fe,Cr)_x(Zr)_y$ was constant during the brazing.



Figure 12. (a) Secondary electron (SE) image of the interface between Cu-7Al-3.5Zr and SS430 in a joint brazed at 1000 °C for 10 min. (b) high-angle annular dark-field (HAADF) image of area X in Figure 12a, and (c) the corresponding elemental spectral analysis for Cu, Al, Fe, Zr, and Cr. It shows a Zr-rich reaction layer and Al-rich sub-layer at the Cu-7Al-3.5Zr and SS430 interface.

Based on the experimental results and above discussion, the formation procedure of the reaction layer in a joint brazed at 1000 $^{\circ}$ C is illustrated with schematics shown in Figure 13. The entire brazing process can be divided into four stages (Figure 13b-e) from the as-cast brazing alloy. Under the solidus temperature, the as-cast brazing alloy was composed of primary α -Cu and eutectic phases $(\alpha$ -Cu + AlCu₂Zr). In the first stage (Figure 13b), when the temperature was raised above the eutectic temperature, where liquid (liquefied eutectic) and solid (α -Cu solid solution) phase of a brazing alloy coexists. Since eutectic phases have an Al- and Zr- rich composition, which lowers melting point of Cu, they were preferentially liquefied. In addition, a load heavier than the light pressure used in common brazing techniques let the liquids phase become tightly wet and evenly spread out onto the α -Al₂O₃. For the second stage (Figure 13c), at 1000 °C, Zr elements in the liquid and decomposed O from α -Al₂O₃ reacted, forming a ZrO₂ reaction layer at the α -Al₂O₃/brazing alloy interface $(3Zr + 2Al_2O_3 = 3ZrO_2 + 4Al)$. The existence of small Cu particles in the reaction layer was developed by the diffusion of Cu in the brazing alloy into the already formed ZrO_2 layer. On the SS430 side, Zr in the brazing alloy was used to form the reaction layer, Zr rich $(Fe,Cr)_x(Zr)_y$, and Al in the brazing alloy and/or excess Al from the reduction of α -Al₂O₃ was used to border the sub-layer—Al-rich $(Fe,Cr)_x(Al)_y$. As the temperature and holding time proceeded during the third stage (Figure 13d), the liquid was used to thicken the ZrO₂ layer following a parabolic late law, and conversely, the volume of the liquefied eutectic near the interface decreased. Finally (as in Figure 13e), since Zr in the liquid phase near the interface was used up in order to make the ZrO₂ reaction layer, inter-diffusion of Al, O and Zr became extinct and the unreacted liquid remained in the brazed seam, which will be solidified as the alternating α -Cu and eutectic phases after cooling.



Above solidus, increasing dwelling time at bonding temperature as liquid-solid state.

Figure 13. Schematic diagram of nucleation and growth processes of reaction layers at the Al₂O₃/Cu-7Al-3.5Zr alloy interface for various time periods.

4. Conclusions

The formation of interfacial reaction layers was investigated in an α -Al₂O₃/430 stainless steel (SS430) joint brazed using a Cu-7Al-3.5Zr active brazing alloy. The specimens were brazed at above its eutectic temperature of 945 °C and below liquidus 1045 °C, where liquid and solid phases of the brazing alloys coexists, and examined in terms of microstructural analysis. The obtained results can be summarized as follows:

(2) The growth kinetics of the layer obey the parabolic rate law with the rate constant of $9.25 \times 10^{-6} \text{ cm} \cdot \text{s}^{-1/2}$. It was observed that a number of low yield strength Cu-rich particles were dispersed over the reaction layer, which can release the residual stress of the joint resulting in a reduction of crack occurrences.

(3) On the SS430 side, Zr in the brazing alloy was used to form the reaction layer, Zr rich $(Fe,Cr)_x(Zr)_y$, Al in the brazing alloy, and/or excess Al from the reduction of α -Al₂O₃, was used to border the sub-layer—Al-rich $(Fe,Cr)_x(Al)_y$.

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