Bonding Mechanism of Electroless Ni-P Film with AlN Substrate and Cu Foil

Chung-Daw Young and Jenq-Gong Duh

Abstract—Electroless Ni (EN) plating method is employed to metallize Y2O3-doped AlN ceramic substrates with different surface morphologies, including the as-received, polished, and etched AlN. The EN-plated AlN substrate is bonded with the Cu foil to form a sandwich-like AlN-EN/Cu/EN-AlN assembly by hot pressing in vacuum with a pressure of 6.5 MPa for 30 min. The bonding strength of the joint is determined by the adhesive abilities of EN/AlN and EN/Cu interfaces. Bonding between the Cu foil and the EN film is achieved by the interdiffusion of Cu and Ni atoms to form a solid solution interlayer, while the EN film is adhered on the AlN substrate mainly by the mechanical interlocking through the rough interface, etched holes, and open pores. At the bonding temperatures lower than 400 °C, no reaction occurs between the Cu foil and the EN film. Increasing bonding temperature enhances the interdiffusion of Cu and Ni to form a strong bonded solid solution interlayer. However, the depletion of the element Ni in EN film results in the formation and enlargement of the pores at EN/AlN interface, which makes a poor adhesion between EN film and AlN substrate. An optimum bonding temperature is observed around 600 and 700 °C for various surface conditions of AlN substrates. Theoretical calculations of the residual thermal stresses in EN, AlN, and Cu layers with respect to the bonding temperatures are related to the variation of the bonding strengths. The relations of the adhesion strengths, interfacial morphologies, elemental distribution, and calculated residual stresses is proposed.

Index Terms—Bonding strength, electroless Ni, interfacial morphology, mechanical interlocking.

I. INTRODUCTION

ALUMINUM NITRIDE (AlN) ceramic with excellent thermal conductivity is considered to be an attractive substrate or packaging material. In addition, AlN ceramics have excellent flexural strength around 5000 kg/cm², high electrical resistivity (10^{14}-10^{24} Ω-cm), low dielectric constant (8.8 at 10 MHz), no toxicity and the coefficient of thermal expansion (4.5 ppm/°C) close to that of Si (4 ppm/°C). These properties make AlN a potential material for application in micromachined packaging as capping/wiring substrates [1] and heat sinks of power devices, such as epitaxial transistors, light emitting diodes and laser diodes [2].

The AlN substrate on which silicon chips are mounted, circuits printed, and connection wires bonded requires surface metallization. Unfortunately, nonwetting of the AlN surface by most pure metals, including Cu, Ag [3], Al [4], and Ni [5] was observed. Nevertheless, AlN reacts with most Ni alloys to form AlNi compounds. Good wetting for Ni alloyed with Ti, Zr, and Hf was reported [5]. A wide variety of processes for metallization of AlN have been proposed, such as thick film metallization [6]–[11]; soldering [12] and brazing [13]; direct bonding [14]; cofiring with high melting point metals [15]; electroless plating [16]–[18]; and hot pressure bonding [19]. Among these metallization methods, the electroless Ni (EN) plating method appears to be the most appropriate one because the thermal expansion coefficient [20(a)] lower than that of Cu and also closer to that of AlN ceramic. Besides, it exhibits the characteristics of excellent solderability, conductivity, receptivity to brazing, wire and die bonding [21], good adhesion [20(b)], well conformity on substrate surface, and interdiffusion with Cu to form a solid solution. Ease of operation, low price and applicability to different kinds of materials, such as metal, plastic and ceramic all make the EN plating method attractive in application.

Although electroless Ni (EN) plating is a promising method for metallizing the oxide ceramics, such as Al₂O₃, few of them concentrate on AlN [16], [17]. It has been reported that the adhesion strength of EN film is related to the surface morphology of AlN substrate. For higher adhesion strength of EN film, the CaC₂-sintered AlN is selectively etched by a NaOH solution in the triple junctions of grain boundaries where Ca atoms are mainly concentrated. Further etching progressively through grains, grain boundaries, and triple junctions enlarges the space between AlN grains, and thus results in a decrease of the adhesion strength. However, the etching behavior of Y₂O₃-sintered AlN substrate [22], [23] and its relation to the adhesion strength of EN film [16], [17], as well as the advanced bondability of the EN-plated AlN substrate with Cu metal are still not yet fully understood.

In this study, solid state bonding of Y₂O₃-sintered AlN and Cu with a deposited EN film as an interlayer was investigated. AlN ceramics of as-received, polished and etched varietier were used in the AlN-EN/Cu/EN-AlN assembly to investigate the effects of surface roughness and morphology on the bonding strength. The thickness of the EN film was varied to evaluate the bondability of the EN-plated AlN substrate.

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and the Cu foil. In addition, the relationships among adhesion strength, elemental redistribution after bonding and estimated residual thermal stresses were also investigated.

II. EXPERIMENTAL PROCEDURES

Electroless Ni plating method was applied on the AlN surface to join AlN ceramic and Cu metal successfully by the hot press method. The AlN ceramic of commercial substrates (SH-15, TOKUYAMA SODA CO., LTD., Japan) with the addition of Y_{2}O_{3} as the sintering aid were used in this study. The specification of the as-received AlN substrates are as follows: size, 1 × 1 × 0.025 in³; surface roughness (Ra), around 0.3 μm; and flatness, ≤0.05/25.4 mm/mm. After SiC grinding, diamond polishing and 4 wt% NaOH solution etching [25], the as-received, polished and etched AlN substrates were prepared for electroless Ni plating.

AlN substrates were surface-pretreated through sensitization and activation by dipping into the acid solutions of 16 g/l SnCl₂ · 2H₂O-30 ml/l HCl and 0.1 g/l PtCl₂-8 ml/l HCl, respectively, for 10 min, and then put into the electroless plating solution for 10 and 60 min, respectively, to have the thin and thick Ni-P films deposited. The constituents in the EN plating bath are as follows: the Ni source is 20 g/l NiSO₄ · 6H₂O, reducing agent is 27 g/l NaH₂PO₄ · H₂O and complexing agent is 16 g/l Na₂H₄C₃O₄ · 6H₂O. The plating solution was adjusted to a pH value of 4.6 by a reaction grade H₂SO₄ acid. Between each step of sensitization, activation, and plating, samples were rinsed in D.I. water for 10 min. A mechanical stirrer was used to agitate and to homogenize the concentration of solution in order to obtain a uniform deposited EN film. The operation temperature was controlled at 70 °C within an isothermal tank.

A Cu foil as an intermediate layer was placed between two EN-metallized AlN substrates to achieve a sandwich-like assembly of AlN/EN/Cu/EN-AlN. Samples were put into a vacuum chamber and heated to the bonding temperatures from 400 to 800 °C. An uniaxial pressure of 6.5 MPa was applied on the samples with graphite punches for 30 min to enhance the intimate contact at the interfaces. Before bonding, the chamber was evacuated to 10⁻⁶ torr. The effects of the different kind of AlN substrates on the bonding strength are compared to each other at 600 °C mainly.

Some of the bonded samples were mounted vertically in epoxy, then ground, polished and carbon coated for interfacial observation. The surface morphologies of the cross-sectional view in the interface and the fractured surface after tensile testing were examined by scanning electron microscopy (SEM, Camscan, U.K.). Backscattered electron image (BEI) was also obtained to distinguish the distribution of EN film from the Cu foil and the AlN substrate. The elemental distribution around the interface and on the fractured surface were measured by electron probe X-ray microanalyzer (EPMA, JXA-733, JEOL, Japan). Structures and phases of the AlN substrate and the electroless Ni plating film were identified by X-ray diffraction (XRD, DMX-B, Rigaku, Japan). Adhesion strength of bonded samples was evaluated by the tensile test, with a pull-off tester (SEBASTIAN FIVE, QUAD, U.S.A.) in which forces were exerted on two pull studs attached on the bonded samples through epoxy, as shown in Fig. 1. In addition, surface roughness of the as-received, polished and etched AlN was measured by a surface profilometer (ALPHA-STEP 250, Tencor, USA).

III. RESULTS AND DISCUSSION

The bonding strength of the sandwich-like AlN/EN/Cu/EN/AlN samples as a function of bonding temperatures has been reported elsewhere [25]: Below 400 °C, no detectable reaction occurs at EN/Cu interface to sustain a sufficient bonding strength. The bonding strength increases initially with increasing temperature above 400 °C, and then decreases gradually through an optimum value around 12 MPa at temperatures between 600 and 700 °C. It is observed that the tensile-tested sample fails at the Cu/EN interface as the sample is bonded below 600 °C. However, the interfacial strength of Cu/EN appears higher than that of AlN/EN as the sample is bonded above 700 °C. Further studies on the bonding behavior and fracture mechanisms are investigated as follows:

A. Effects of AlN Surface Morphology on the Bonding Strength of AlN/EN and Cu/EN Interfaces

AlN substrates were untreated, polished, and etched to achieve different surface morphologies for the electroless Ni (EN) plating. The EN-plated AlN substrates were bonded with the Cu foil to form a sandwich-like AlN/EN/Cu/EN-AlN assembly by hot pressing in vacuum. An optimum bonding temperature is selected at 600 °C [25]. The thickness of the EN film is around 1 μm. Typical cross-sectional view of the bonded samples with the previously etched AlN, and the corresponding BEI are shown in Fig. 2(a) and (b), respectively. The white grains in the dark AlN matrix, named Al-Y-O compounds, are identified in structure by XRD [6] and distinguished qualitatively from the AlN grains in composition by EPMA. The penetration of the EN film into the AlN substrate is clearly observed around the Al-Y-O compounds at the AlN/EN interface, while the interface between the Cu foil and the EN film is not visible. At the EN/AlN interface, the electroless Ni-P precipitating at the etched AlN/Al-Y-O grain boundaries to form the anchor-like interlocking is thought to be the major bonding mechanism in joining the EN film and the AlN substrate. A similar case for the interlocking between the EN film and the AlN substrate, and the related surface mor-
Fig. 2. (a) Cross-sectional view of the bonded sample containing etched AlN and 1 μm-thin EN interlayer and (b) corresponding BEI of (a), showing the distribution of the EN film, the Al-Y-O compounds, and the AlN matrix.

Fig. 3. Unbonded regions observed in the Cu/EN interface using the as-received AlN substrate metallized by EN and bonded to Cu.

Fig. 4(a) and (b). Isolated EN fragments with irregular shape are observed on the AlN fracture surface. Small amount of area of the fracture surface is occupied by the residual EN films. It is suggested that the fracture of the bonded samples is basically along the AlN/EN interface, yet partially through the EN/Cu interface to form the residual EN films. Most of the residual EN films surround the Al-Y-O compounds, as shown in Fig. 4(c). The corresponding X-ray mapping for Ni is indicated in Fig. 4(d). The reason for forming the residual EN films is attributed to the fact that the EN film adheres strongly on the AlN surface through the anchor-like interlocking with the holes etched around the Al-Y-O compounds. The conformal EN film exhibits concave shape at these corresponding sites and is accompanied with the formation of noncontact area at the Cu/EN interface. Therefore, cracks advancing along the AlN/EN interface will change their propagation direction easily after being obstructed by the EN anchors. These anchors will be cut through so that the cracks will propagate along the Cu/EN interface temporarily until they return back to the AlN/EN interface to meet the other cracks. A simplified model to show the cross-sectional view of the Cu/EN/AlN structure and its fracturing mechanism is schematically presented in Fig. 5(a).

On the contrary, the residual EN films adhering on the fracture surface of the untreated AlN substrate are not around the Al-Y-O compounds, as shown in Fig. 6. This is because...
no EN anchors formed at the grain boundaries around the Al-Y-O compounds on the untreated AlN surface. Furthermore, for the polished AlN substrates the residual EN films only reside around the open pores on the fractured AlN surface, as indicated in Fig. 7. The open pores, which are produced at the sintering stage in manufacturing the AlN substrate, provide another kind of anchor sites for the bonding. This kind of anchor site is unexpected and the density of this type of anchor site is not sufficient to achieve good bonding with the EN film. As a comparison, the untreated AlN substrate with identical surface roughness as the etched is rougher than the polished, and thus has higher adhesive ability for the EN film. Because of the difference between the surface morphologies of the AlN substrates, the average bonding strengths for the 600 °C-bonded sandwich-like samples with the AlN substrates etched, untreated and polished are 13, 10, and 6 MPa, respectively.

B. Thickness Effects of EN Films on the Bonding Strength of AlN/EN and Cu/EN Interfaces

The etched AlN substrates were metallized by the electroless Ni plating up to 120 min to obtain a thick EN film, and then bonded with the Cu foil at 600 °C to form the sandwich-like assembly. The cross-sectional view of the sample is shown in Fig. 8(a). The thick EN interlayer can be distinguished from the AlN substrate and the Cu foil easily through the BEI, as shown in Fig. 8(b). The thickness of the EN interlayer is estimated around 10 μm. The anchor-like interlocking at the AlN/EN interface are observed in Fig. 8(c). The thick EN interlayer appears gray in color, as indicated in BEI in Fig. 8(d). A close look of the image exhibits that some white, small and isolated particles distribute inside the EN film, as shown in Fig. 8(e). The particle size is smaller than submicron. This implies that the thick EN film consists of two phases. By XRD analysis, the fracture surface of the EN film at the AlN/EN interface shows two types of structures, i.e., Ni and Ni₃P, as indicated in Fig. 9. Because the atomic number of the element P is smaller than that of Ni, the BEI of the Ni₃P will appears darker than that of Ni. The Ni-P phase of the EN interlayer transform from its original amorphous state into crystals after 600 °C bonding, which makes the isolated Ni micrograins distribute inside the Ni₃P matrix. The

Fig. 4. Fractured surface of etched AlN substrate bonded at 600 °C for 30 min after the tensile testing. (a) Scanning electron microscope (SEM) image showing the surface morphology, (b) corresponding BEI of (a), (c) higher magnification pictures of (b) showing the residual EN film-surrounded Al-Y-O compounds, and (d) corresponding X-ray map for Ni Kα of (c).
phase transformation affected by the elemental diffusion will be explained by the following.

At the Cu/EN interface, the bonding between the Cu foil and the EN film is achieved through the interdiffusion of Cu and Ni. The cross-sectional view and the elemental distribution for elements P, Ni, Cu, and Al in the Cu/EN interface are shown in Fig. 10(a)–(e), respectively. The distribution of element P reflects the shape of the EN film. Ni, however, distributes across the Cu/EN interface and partially overlaps with the Cu foil. A transition region is thus formed with the Ni-rich side near the Cu/EN interface and the Cu-rich side adjacent to the Cu foil. Interdiffusion of Cu and Ni [27] results in the formation of a (Cu, Ni) solid solution layer. Although the diffusion rate of Cu in Ni is faster than that of Ni in Cu [28], the diffusion barrier effect of EN [29] could slow down the diffusion of Cu into the Ni-Ni₃P binary system. Therefore, diffusion of Ni into the Cu foil could make Ni depleted inside the EN film and result in precipitation of small Ni grains in the Ni₃P matrix. In fact, the Ni₃P microcrystals will precipitate in the Ni matrix after 600 °C heat treatment [30] if no mass transport disturbs the composition of this film. It should be noted that the (Cu, Ni) solid solution layer is not yet found in the thin EN film-contained samples in SEI under the same magnified scale. Due to the formation of a thick (Cu, Ni) solid solution layer, the joint strength at the Cu/EN interface is supposed to be stronger for the thick EN film samples than the thin ones. The resultant bonding strength (16.39 ± 1.99 MPa) for the 10 μm-thick EN-interlayered bonded samples is therefore higher than that of the 1 μm-thin one (12.07 ± 5.25 MPa). Furthermore, the bondability of the thicker EN film with the Cu foil results in less residual EN films on the AlN fracture surface, as shown in Fig. 11(a).

Both the thick and thin EN film bonded samples have the anchor-like interlocking at the EN/AlN interface due to the
etched AlN surface. However, morphologies of the fracture surfaces on the AlN sides are different. For the thick EN film case, fewer residual EN on the AlN fracture surface is observed, as shown in Fig. 11(a). The residual EN films only exist at the grain boundaries around the Al-Y-O compounds and appear in circular shape [Fig. 11(b)]. It seems that the fracture takes place only by crack propagating along the AlN/EN interface and cutting across the EN anchors. There are no large amounts of EN films left on the AlN surface and no fracture occurred at the Cu/EN interface. The thick EN film covers and fills up the etching holes on AlN surface and makes contacts on the other side with the Cu foil rather easily due to the gently wavy surface [26] such that it tends to form a good EN/Cu joint with Cu. For the thin EN film case, the EN film contours the etched AlN surface and performs a similar asperity making good Cu/EN contact difficult. This implies that failure may also partially occur through the Cu/EN interface, especially near the EN anchors, in addition to those mainly through the AlN/EN interface. In fact, time is required for Cu to diffuse and deform to fill up the concave holes on the thin EN film surface if the surface is rough. The proposed models to elucidate the different bonding mechanisms of the Cu/EN/AlN assembly for thin and thick EN film are schematically presented in Fig. 5(a) and (b).

C. Effect of Phase Transformation of the EN Film on the Bonding Strength

The morphology of the fracture surface for the EN film on the Cu foil as failure goes through the EN/AlN interface is shown in Fig. 12(a). Basically, it is a replica of the AlN surface. From the corresponding BEI [Fig. 12(b)], small pitting holes are homogeneously distributed in the EN replica. To evaluate the correlation between the pitting holes and the bonding temperatures, the polished AlN substrates plated with the EN films and then bonded with the Cu foil are studied. The pitting holes in the EN replicas from the samples bonded at 150, 600, and 700 °C are shown in Fig. 13(a)–(c), respectively. The sizes of the pitting holes increase with the increasing bonding temperatures. On the basis of the Griffith theory, a specimen with larger hole size has lower average applied strength at which the crack is located. Thus, the bonding strength will decrease as the pitting holes develop and become
enlarged. In addition, cracks are found in the EN replica, as shown in Fig. 12(b). This implies that cracks can propagate more easily along the AlN/EN interface as compared with the Cu/EN interface. The reason is not only due to the strong bonding at the Cu/EN interface by forming a (Cu, Ni) solid solution layer, but also due to the pitting holes at the AlN/EN interface. Small particles extruding out of the residual EN film toward Cu side at the fractured Cu/EN interface appear on the sample
Fig. 9. Phases of the EN films under different heat treatments. (a) as-plated and (b) bonded at 600 °C for 30 min.

edges, as shown in Fig. 14(a). Accompanied with the results of XRD analysis (Fig. 9), the light image in BEI indicates that the extruded particles should be Ni, while the dark base is the Ni₃P matrix [Fig. 14(b)]. For the as-plated amorphous EN film containing high phosphorus content of 11.3%, phase transform takes places at the temperatures higher than 400 °C, and results in embedded Ni particles inside the Ni₃P matrix. It is interesting to note that the extruded Ni particle and the small pitting hole have the identical size smaller than 500 nm, which is also the size of the Ni particle distributed inside the EN film. Due to the continuous solid solution reaction between the EN film and the Cu foil, Ni is depleted in the EN film. Thus, Ni becomes the isolated particle embedded in the Ni₃P matrix. Thereafter, nickel diffuses from the AlN/EN interface, through the Ni₃P matrix, into the (Cu, Ni) solid solution layer. This induces Ni extrusion from the EN film at the Cu/EN interface and the concave interfacial holes on the other side. A simplified model to represent the interfacial reaction of the Cu/EN/AlN structure as described above is schematically shown in Fig. 15. Although the elevated bonding temperatures enhance the joining between the Cu foil and the EN film, the formation of the pitting holes in the AlN/EN interface creates additional sites and paths for the cracks initiation and propagation, which will possibly result in weak bonding for a long period of time. It is believed that, in addition to thermal stress effects, the pitting holes cause the decrease of bond strength as the bonding temperature increases above the optimum temperature between 600 and 700 °C [25].

D. Estimation of the Residual Thermal Stress in the Multilayered Structure

It is difficult to estimate the residual thermal stress in the bonded metal/ceramic multilayer structure because of the insufficient information concerning the material properties at various temperatures and, as well as the lack of information on the boundary conditions on both the free surface and the interface. However, if the discontinuity at the interface is ignored, i.e., only the average residual stress of each layer subjected is considered, the residual stress can be estimated. The approach used to calculate the stresses is indicated in Fig. 16. The EN film is plated on the AlN substrate and constrained each other during a heat treatment process from the plating temperature through the bonding temperatures until cooling down to the room temperature. Because the stacked multilayered samples are loaded axially in the chamber at room temperature before bonding, the slightly bent EN-plated AlN samples are expected to flatten. Therefore, at the as-plated stage, the thermal stresses in AlN ceramic, σₖᵣ, and in EN interlayer, σₖ₁, can be evaluated by the equilibrium of forces and the uniformity of strain as [31]

$$\sigma_c^a \cdot t_c + \sigma_i^a \cdot t_i = 0$$

$$\frac{\sigma_c^a \cdot (1 - \nu_c)}{E_c} + \alpha_c \cdot \Delta T^a = \frac{\sigma_i^a \cdot (1 - \nu_i)}{E_i} + \alpha_i \cdot \Delta T^a$$

where the subscripts c and i signify the AlN ceramic and the EN film, respectively, and the superscript a means that the sample is in the as-plated stage. t, ν, E, and α represent the thickness, Poisson’s ratio, Young’s modules and the coefficient of thermal expansion of each material components in the EN-plated samples, and ΔT is the temperature change from the EN plating temperature to the room temperature. Similarly, at the heating stage, these two equilibria and constraint conditions are also satisfied. Thus, one obtains

$$\sigma_c^b \cdot t_c + \sigma_i^b \cdot t_i = 0$$

$$\frac{\sigma_c^b \cdot (1 - \nu_c)}{E_c} + \alpha_c \cdot \Delta T^b = \frac{\sigma_i^b \cdot (1 - \nu_i)}{E_i} + \alpha_i \cdot \Delta T^b$$

$$\frac{\sigma_c^b \cdot (1 - \nu_c)}{E_c} + \alpha_c \cdot \Delta T^b = \frac{\sigma_i^b \cdot (1 - \nu_i)}{E_i} + \alpha_i \cdot \Delta T^b$$

$$\frac{\sigma_c^b \cdot (1 - \nu_c)}{E_c} + \alpha_c \cdot \Delta T^b = \frac{\sigma_i^b \cdot (1 - \nu_i)}{E_i} + \alpha_i \cdot \Delta T^b$$

$$\frac{\sigma_c^b \cdot (1 - \nu_c)}{E_c} + \alpha_c \cdot \Delta T^b = \frac{\sigma_i^b \cdot (1 - \nu_i)}{E_i} + \alpha_i \cdot \Delta T^b$$
where the superscript \( h \) means that the sample is heated up to the bonding temperature and \( \Delta T^h \) is the temperature change from room temperature to the bonding temperature. During the cooling stage one Cu foil and two EN-plated AlN samples are bonded together and constrained each other. There is no bending of the bonded samples after removing them from the chamber because of the symmetry in the multilayered structure. The equilibrium of force and the uniformity of strain
Fig. 11. Fractured surface of the etched AlN substrate bonded at 600 °C for 30 min with a 10-μm thick EN film as the interlayer. (a) SEM image, (b) corresponding BEI showing the distribution of the residual EN film and the Al-Y-O compounds.

Fig. 12. Fracture surface morphology of the EN film on the Cu foil as fracture through the EN/AlN interface. (a) SEM image and (b) corresponding BEI.

The results of the estimated residual thermal stresses are shown in Fig. 17. At the AlN/EN interface the interfacial residual thermal stress \( \Delta \sigma_c \) are nearly a constant around 32.5 MPa independent of present bonding temperatures, while at the Cu/EN interface the interfacial residual thermal stress \( \Delta \sigma_m \) varies from 26.5–31.5 MPa as the bonding temperature ranges from 400 to 800 °C. Additionally, there is no appreciable difference in the calculated residual thermal stresses between the samples containing 1 and 10-μm thick EN films, respectively.

E. Optimum Bonding Strength

At bonding temperatures below 400 °C, no significant reaction, and thus no effective bonding strength is achieved.
between the Cu foil and the EN film. Increasing bonding temperature enhances the interdiffusion of Cu and Ni to form a solid solution interlayer to achieve a strong bonding. The bonding strength accordingly increases with the bonding temperatures below 600 °C, as shown in the right down corner of Fig. 18. However, at elevated temperatures the depletion of element Ni in EN film to form a (Ni, Cu) solid solution layer simultaneously results in the formation and enlargement of pores at the EN/AlN interface, which unfortunately makes a poor adhesion of EN film on AlN substrate. The bonding strength therefore decreases due to the degradation of AlN/EN interface as the bonding temperature is further raised above 700 °C. Finally, an optimum bonding strength ranging from 6 to 16 MPa occurs between 600–700 °C for various surface states of AlN substrates used.

Surface roughness of AlN is another important factor affecting the bonding strength in the joint. As the AlN substrate is etched and the surface roughness increases, the mechanical interlocking between EN film and AlN substrate around the etching holes enhances the strength of AlN/EN interface. A summary for bonded samples with as-received, polished, and etched AlN, designated by symbols A, P, and E, respectively, is shown in Fig. 18. A further increase in bonding strength is also achieved by thickening the EN film. The advantage of thick EN film is not only by forming a thick (Ni, Cu) solid solution reaction layer at the EN/Cu interface, but also by covering the rough surface of etched AlN substrate to assure the easy contact with Cu foil. The bonding strength of the thick EN-contained sample, designated by symbol T in Fig. 18, is therefore higher than that of the thin one, i.e., the sample E. The optimum condition to achieve highest bonding strength in the sandwich-like AlN/EN/Cu/EN/AlN bonded sample in the study is: the AlN surface is etched, the EN film is plated thicker and the bonding temperature is between 600 and 700 °C.

The “measured” bonding strength as mentioned above is the mechanical strength of a joint under a residual thermal stress, which is induced by either the mismatch of thermal expansion coefficient of the joined materials or the temperature difference they experience. This implies that the bonding strength could be increased by the amount of the residual thermal stress if the thermal stress is removed. The bonding strength of a joint without residual thermal stresses is regarded as the
“intrinsic” bonding strength. The intrinsic value is attributed to interfacial strength of both AlN/EN and EN/Cu. If the bonding temperature is below the optimum value, the intrinsic bonding strength is the sum of the measured bonding strength and the interfacial residual thermal stress, $\Delta \sigma^{\text{irr}}$. This is presented as the “intrinsic bonding strength of Cu/EN” in

**Fig. 14.** Residual EN films on AlN fracture surfaces near the edges of the bonded sample. (a) SEM image. (b) Corresponding BEI showing two phases inside the heat-treated EN film.

**Fig. 15.** A simplified model to show the bonding mechanism and the interfacial reaction of the Cu/EN/AlN structure.

**Fig. 18.** The strength increasing with temperature is explained by the formation and growth of the (Cu, Ni) solid solution layer at the Cu/EN interface. On the contrary, the intrinsic bonding strength of AlN/EN at temperatures higher than the optimum one is equal to the sum of the measured bonding strength of joint and the residual thermal stress at the AlN/EN interface, $\Delta \sigma^{\text{irr}}$. The sum of measured adhesion strength and residual thermal stress of the as-plated EN/AlN interface is related to the room temperature one. Combined with the two groups of intrinsic values, the trend of “intrinsic bonding strength of AlN/EN” is traced in Fig. 18. No significant increase of the intrinsic strength with temperature is observed initially. The small increment of strength is supposed to be due to the change of the EN film in the mechanical properties, such as the hardening, after heat treating. The decrease of strength with temperature above 700 °C is explained by the formation of pitting holes at the AlN/EN interface. There are two critical points in the figure. At first, the strength of a joint is dependent on both interfaces of AlN/EN and EN/Cu, and is controlled by the weaker one. Second, there is still space for further improving the bonding strength up to the intrinsic value by removing the residual thermal stress from the joint. It should be pointed out that Fig. 18 only explain the relations between intrinsic and actually measured bonding strength of the joints qualitatively. However, this approach provides a guide to systematize the design of the joint for commercial

**TABLE I**

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<th></th>
<th>Young’s modulus (GPa)</th>
<th>Thickness (nm)</th>
<th>Thermal expansion coefficient ($\mu$m/m°C)</th>
<th>Poisson’s ratio</th>
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<td>17</td>
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<td>50</td>
<td>500</td>
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<td>12</td>
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Mechanical Properties Used in Calculating the Residual Thermal Stress in the AlN/EN/Cu/EN/AlN Bonded Samples
and industrial applications, if the intrinsic (electronic, atomic, and/or mechanical) bonding strength and the distribution of the residual thermal stress around the joint could be estimated precisely.

IV. CONCLUSION

1) Yttrium-containing AlN ceramic substrates were metallized by electroless Ni (EN) plating and bonded with a Cu foil by hot pressing in vacuum to form a multilayered AlN-EN/Cu/EN-AlN structure. The bonding strength of the samples was measured by tensile test method.

2) Joining of AlN and EN is through the mechanical interlocking, as the EN anchors penetrate into the etching sites around the Al-Y-O compounds. However, the bonding between Cu and EN is achieved by the chemical reaction to form a (Cu, Ni) solid solution interlayer.

3) Formation of (Cu, Ni) solid solution interlayer is accompanied by depletion of Ni in the EN film. This enhances the phase transformation of EN film from an amorphous Ni-P single phase into a binary phase, which is composed of Ni microcrystals embedded inside the Ni$_3$P matrix, and thus leaves pitting holes in the EN film along the EN/AlN interface.
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4) Elevated bonding temperatures promote the joining of Cu and EN by enhancing the elemental diffusion and solid solution reaction, while formation and enlargement of pitting holes in the AlN/EN interface deteriorate the bonding strength. An optimum bonding strength occurs as the contributions of the adhesion strengths at AlN/EN and Cu/EN interfaces are competitive.

5) Samples with 10-μm thick EN films exhibit higher bonding strength than those with 1 μm-thin ones because of the good contact and thick solid solution interlayer formed between the Cu foil and the EN film. The result is that cracks propagate only along the EN/AlN interface and cut through the EN anchors.

6) Intrinsic (electronic, atomic and/or mechanical) bonding could be qualitatively related to the measured bonding strength by calculation of residual thermal stresses. This approach provides a guide to systematize the joint design for commercial and industrial applications.

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