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High temperature reliability of power module substrates

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Abstract

The thermal cycling reliability of candidate copper and aluminium power substrates has been assessed for use at temperatures exceeding 300°C peak using a combination of thermal cycling, nanoindentation and finite element modelling to understand the relative stresses and evolution of the mechanical properties. The results include the relative cycling lifetimes up to 350°C, demonstrating almost an order of magnitude higher lifetime for active metal brazed Al / AlN substrates over Cu / Si_3N_4, but four times more severe roughening and cracking of the Ni-P plating's on the Al / AlN (DBA) substrates. The nonlinear finite element modelling illustrated that the yield strength of the metal and the thickness of the ceramic are the main stress controlling factors, but comparisons with the cycling lifetime results demonstrated that the fracture toughness (resistance) of the ceramic is the over-riding controlling factor for the overall passive thermal cycling lifetimes. In order to achieve the highest substrate lifetime for the highly stressed high temperature thermal cycled applications, the optimum solution appears to be annealed copper, brazed on to a thicker than normal or higher fracture toughness Si_3N_4 ceramic.

1. Introduction

One of the main barriers to high temperature operation of wide bandgap power semiconductor devices above 200°C is current power module packaging technology. Many of the materials used currently for encapsulation, moldings or solders for example will degrade or creep very quickly above this temperature. However, other materials used currently below this temperature might be candidates for use at higher temperatures. Such materials include direct-bonded copper (DBC) type substrates, including aluminium and active metal brazed (AMB) variants. Aluminium based ceramic substrates are often generally known as direct-bonded aluminium (DBA) substrates, even though they may actually be active metal brazed.

Other investigators have already demonstrated the reliability and some of the failure mechanisms of these types of substrates at lower temperatures. The copper DBC and brazed type substrates and overview of manufacturing conditions were described more than 15 years ago [1], indicating that these processes are conducted at around 1000°C. Tapering the edges of the metal was shown to increase the reliability of these types of power substrates. The same edge tapering technique has also been demonstrated up to a peak cycling temperature of 250°C [2]. The advantages and disadvantages and properties of the different ceramic materials have also been compared and contrasted more recently [3]. The reliability of silver sinter die attach and DBA thermal cycling performance has also been demonstrated before at lower temperatures than studied in the present work [4]. The present author has already disclosed some results for passive thermal cycling up to 350°C [5].

In this work, we continue to investigate the reliability of all available types of aluminium and copper substrate materials manufactured with Alumina (Al_2O_3), aluminium nitride (AlN) and the more recently available and more expensive silicon nitride (Si_3N_4) ceramics for an automotive drive application that has a constantly varying load, a peak temperature exceeding 300°C and a steady-state ambient temperature during operation approaching

100°C (lower at t = 0). Our overall goal has been to evaluate and test compatible combinations of the most reliable materials such as those found in this work, including power semiconductors, wire bonds and die attach technologies at temperatures over 300°C, and some of these results have already been published elsewhere [6,7], including for DBCs used at high frequencies [8].

2. Experimental Method

Substrates were placed vertically in slotted stainless steel card racks mounted to stainless steel wire mesh shelves in a purpose made passive thermal cycling oven with resistive heating and fan assisted liquid nitrogen cooling. The substrates were rotated periodically in the vertical plane (within the slots) and the shelves were swapped and rotated periodically in the horizontal plane to equalize the thermal conditions seen by each material as far as practical. The substrates were removed and inspected under optical microscope for signs of damage every 10 cycles initially, and materials that demonstrated exceptional lifetimes were removed and inspected every 30 to 50 cycles. Two thermal cycling profiles were used: 100-350°C and 0-350°C with ramp times of approximately 50°C per minute and peak hold times of around 3 minutes. Other samples, including die attach test structures and full assemblies with silicon carbide dies and baseplate were tested at the same time and inspected in the same way. Nanoindentation measurements were carried out on the surface of the substrate materials before and periodically during thermal cycling to measure the evolution of mechanical properties versus the number of thermal cycles. An optical inferometer was used to measure the surface roughness for some materials.

3. Experimental Results

The passive thermal cycling lifetime results for the two deep thermal profiles are listed in table-1 along with the main substrate specifications. From these results, it is clear that the aluminium-aluminium nitride (Al / AlN) AMB substrates have the highest lifetimes, at least seven times that of the next best substrate type, copper-silicon nitride (Cu / Si_3N_4) AMB. All substrate types except for the silicon nitride DBC were declared failed at the first obvious signs of concoidal fracture of the ceramic, which can be seen clearly under optical microscope. The silicon nitride DBC failed at the bond itself with no concoidal fractures present; these results also demonstrate that the AMB bond is stronger than the DBC bond (at least) on silicon nitride ceramics, validating other work [1].

Table-1 Measured thermal cycling lifetimes of bare substrates

Ceramic	**Bond type**	**Conductor**	**Plating**	**100-350°C lifetime (cycles)**	**0-350°C lifetime (cycles)**
AlN, 0.6 mm	DCB	Cu, 0.3 mm	ENIG	50	12
Al_2O_3, 0.4mm	DCB	Cu, 0.3 mm	Ni-P	75	13
Si_3N_4, 0.3mm	DCB	Cu, 0.3 mm	None	1160	124
Si_3N_4, 0.3mm	AMB	Cu, 0.5 mm	None	1100	145
Si_3N_4, 0.3mm	AMB	Cu, 0.5 mm	ENIG	-	450*
AlN, 0.6 mm	AMB	Al, 0.4 mm	Ni-P	>>1300**	>>990**

*Bonded to thick baseplate, with silver sintered silicon carbide dies and aluminium wire bonds.

**No metal-ceramic interface delamination or ceramic fractures to date. Tests still on-going.

The relative rates of substrate metal surface roughening and cracking, including of electroless nickel (EN) plating with phosphorous contents of approximately 10% and Ni-P plating's with a final immersion gold (ENIG) coating have been monitored in this environment. Other authors have demonstrated and quantified these effects for cycling temperatures up to 250°C, showing a peak to peak roughness of 10 µm, and attributed the roughnening to aluminium reconstruction and grain boundary sliding [9]. Our results also showed a trend towards early cracking of the plating's on Cu /Si_3N_4 substrates, at around ten 0-350°C cycles, followed by only a small amount of additional roughening by end of life. However, the plated Al / AlN variants show hillocks or ridges forming first, as shown in fig-1, followed later by plating cracks after approximately 100 cycles that become progressively more severe up to a peak-to-peak roughness of approximately 60 µm, as shown in fig-2, compared to 15 µm for ENIG plated copper AMB substrates. There have been no aluminium-ceramic failures to date (these tests are on-going).

Our other results have shown that the same Cu / Si_3N_4 AMB ENIG plated silicon nitride substrates fixed to a thick baseplate survived to approximately 450 cycles before concoidal fractures occurred, approximately three times more cycles than without a baseplate. In this case, the plating cracks and surface roughening were shown not to be detrimental to either silicon carbide dies attached with silver sinter paste, or the large and small diameter aluminium wire bonds.

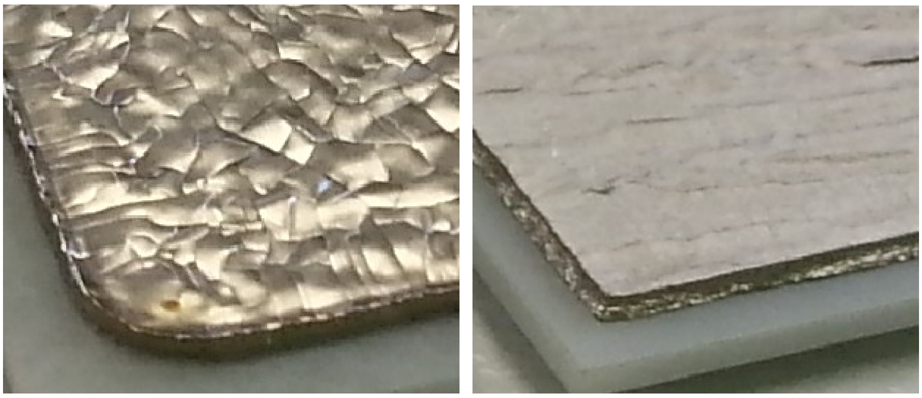

Fig. 1. Magnified views of two different Ni-P plated substrates Al / AlN substrates before and after thermal cycling clearly indicating hillocks / ridges and cracking of the plating's.

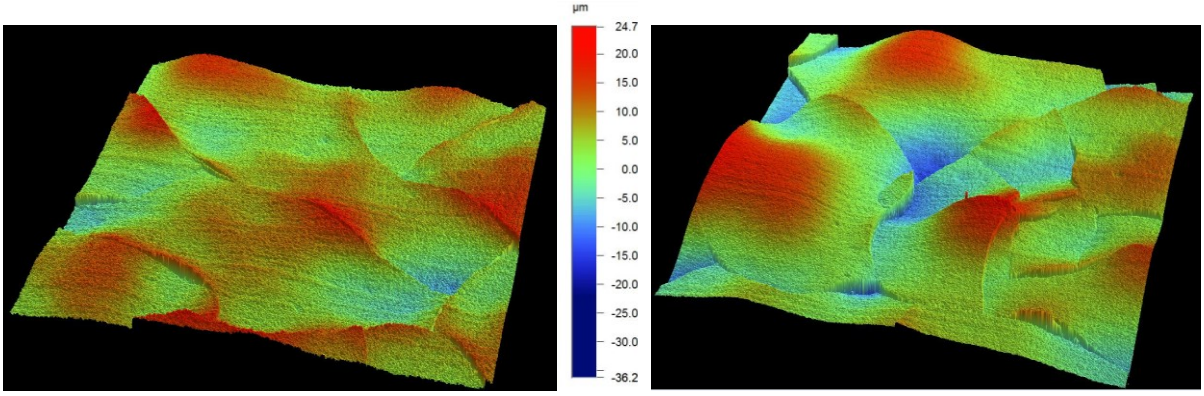

Fig. 2. Optical inferometer images of Ni-P plated aluminium substrate 2mm x 2mm surfaces after 65 cycles (left) and 200 cycles (right) thermal cycles of 0-350°C, illustrating the plating crack evolution.

4. Finite Element Modelling Results

Thermal stress simulations have also been carried using a hypothetical structured substrate design to simulate the relative stresses imposed by the metal on the ceramic surface, the cause of the concoidal fractures witnessed in thermal cycling tests, for copper and aluminium-ceramic substrates, both with and without baseplates. These results, shown in fig-3, indicate that the thermal stresses are approximately 45 % higher when mounted to a thick baseplate than the situation when the substrates are not mounted to a baseplate (floating) and therefore completely free to deform when heated and cooled in the cycling oven.

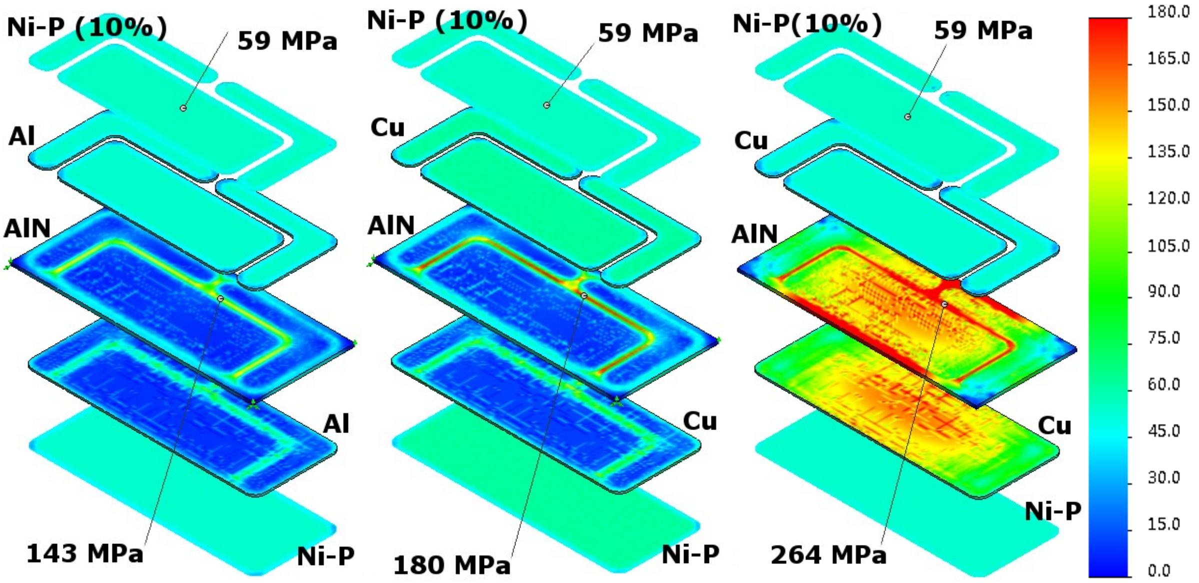

Fig. 3. Exploded views showing simulated Von Mises stresses for Ni-P plated DBC and DBA substrates at 300°C: Cu/AlN/Cu floating (left), Al/AlN/Al (centre) and Al/AlN/Al on a rigid baseplate (right).

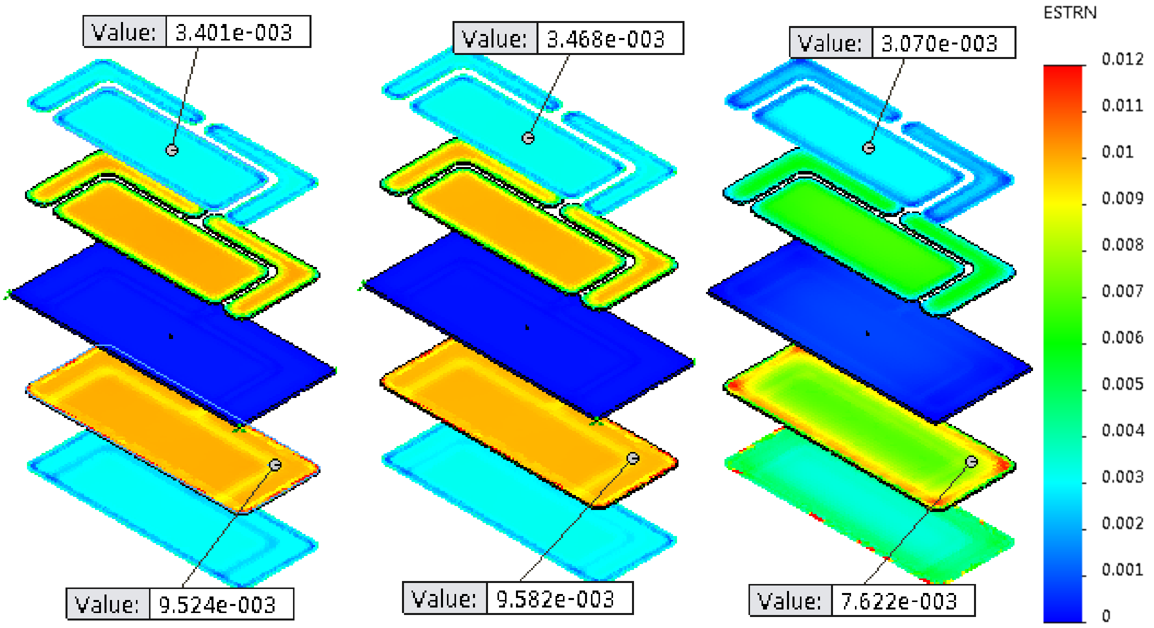

Fig. 4. Exploded views showing simulated equivalent strains for Ni-P plated DBC and DBA substrates at 300°C: Cu/AlN/Cu floating (left), Al/AlN/Al (centre) and Al/AlN/Al on a rigid baseplate (right).

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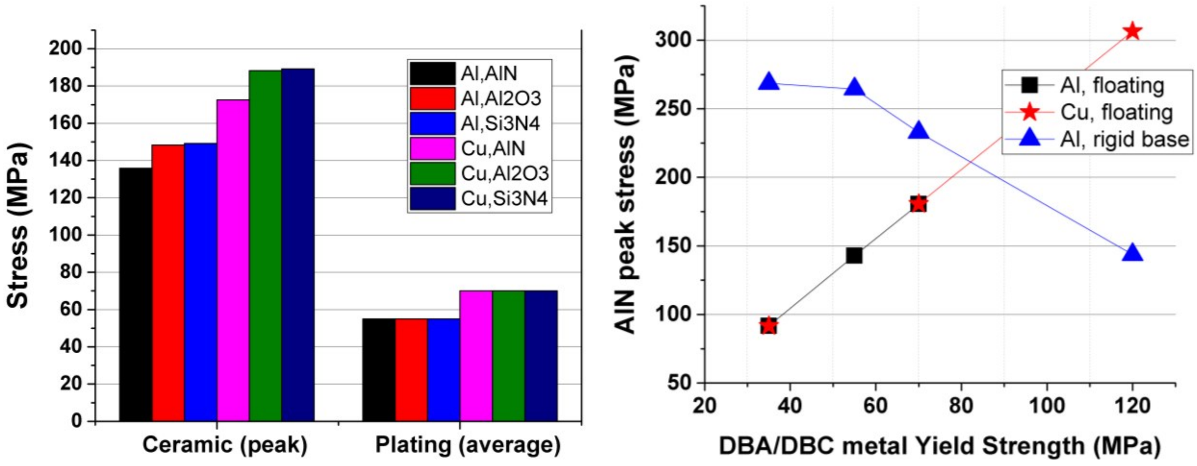

Fig. 5. Column plot shows the modelled peak stress on the ceramic surface at the metal-ceramic interface and average Ni-P plating stress for DBA and DBC substrates with aluminium nitride, Alumina and silicon nitride ceramics. Scatter plot shows the modelled peak stress level on the ceramic surface at the metal-ceramic interface for DBA and DBC substrates using aluminium nitride, including the scenario where the DBA is fixed to a rigid baseplate.

4.1. Stress and strain dependence of Ni-P plating properties

The dependence on the Ni-P plating properties has been modelled. The elastic modulus of these electro-less coatings is known to be dependent on its phosphorous (P) content, but only varies within a small range. However, heat treatment of these coatings is known to increase both the hardness and modulus significantly and this was also found to be the case for our Ni-P plated thermal cycled substrates; after just one hundred 0-350degC cycles, the modulus increased from approximately 120 GPa to 170 GPa. The range of Ni-P elastic modulus from 110 to 220 GPa was modelled and the results showed that the differences in stress levels and distribution at all layers of the DBA substrate due to the Ni-P plating mechanical properties were negligible, presumably due to the Ni-P coating being so thin (5 microns thick). The modelled strains, shown in fig-4, indicate that the strains are similarly low for both copper and aluminium, which implies that the strain within the plating is not contributing significantly to the Ni-P plating cracks.

4.2. Stress dependence of ceramic type

The DBA and active metal brazed silicon nitride substrates appear to have the high thermal cycling lifetimes at high peak temperatures. The dependence on the ceramic type (its mechanical properties) with identical metal and ceramic layer thicknesses, was modelled. The results are shown in fig-5 column plot and an inspection of the mechanical properties of each ceramic material indicate that the peak stress has a small dependence on the elastic modulus, with that of the aluminium nitride material being slightly higher at 330 GPa than for the Alumina and silicon nitride materials, both 303 GPa.

4.3. Stress dependence of metal type

The metal properties have a much greater influence on the stress levels than the ceramic properties; an inspection of the metal properties indicated that the yield strength value of the metals is the dominant property, as indicated by the scatter plot of fig-5. This property is dependent on the state of anneal of the metal following the substrate manufacturing processes. These values are not generally available from the manufacturers but the practical range can be obtained from available literature for modelling purposes. We presume that, due to the extremely high temperature substrate manufacturing processes, the metals are in a relatively high state of anneal as received (before thermal cycling). Our attempts to measure the properties of the copper using the nanoindentation technique have proven inconclusive to date, but it is expected that the actual values for the aluminium based materials will be somewhere in the range of 35 to 80 MPa and 50 to 100 MPa for the copper based substrates. Judging from the nanoindentation hardness values obtained so far before and after passive thermal cycling, there is little or no work hardening and therefore no change in yield strength during cycling. The reason for this is perhaps because the peak cycling temperature of 350°C used in this present work is high enough to keep the metals in an annealed state, to counter any increase in work hardening that might occur due to regular deformation of the substrates during cycling.

4.4. Stress dependence on mounting type

One of the goals of this modelling work was to determine the conditions that the substrate materials would see in the passive cycling environment compared to an active cycling environment, with and without bonded baseplates. As such, fig-5 also illustrates the yield strength dependence for a DBA substrate mounted on to a rigid (immoveable) base. This result demonstrates an opposite dependence on the yield strength to that of the DBA and DBC models that are un-mounted and therefore allowed to bend and deform as they would in the passive cycling oven.

4.5. Stress dependence of commercially available substrates

The most common commercially available combinations of aluminium (DBA) and copper thicknesses (AMB), and ceramic thicknesses, were also modelled to determine their relative peak stresses on the surface of the ceramics. These were the exact same thickness combinations as the ones that were passive thermal cycled. These modelling results demonstrated that the practical metal thicknesses from 0.3 mm to 0.5 mm have almost no influence on the stress levels seen on the ceramic surfaces, or the Ni-P platings. However, the much thinner silicon nitride ceramic does show a marked increase in the peak stress level.

5. Conclusions

The relative cycling lifetimes of various power module substrates and a copper-silicon nitride AMB based power module have been determined experimentally for a very high temperature thermal cycled application, for two representative thermal cycling profiles. These experimental results demonstrated the superior lifetime of Al / AlN AMB substrates compared to Cu / Si_3N_4, but progressively severe roughness and cracking of the former, approximately four times greater than for the copper substrates. Since the experimental results have showed that most of the substrates failed by concoidal fracture of the ceramic where it interfaces to the bonded metal sheets due to the imposed thermo-mechanical stresses, finite element modelling was carried out to determine the relative stresses of the different metal and ceramic material sets in use. A representative hypothetical substrate model was used in order to determine the relative levels of stress at those interfaces and to model the influences of the thermal and mechanical properties of the substrate materials, including the final plating. These results may prove useful when selecting substrate materials for thermally cycled applications with high peak temperatures. It was found that the main influence on the peak stress seen at the metal-ceramic interface, at least for the passive thermal cycling scenario, is the yield stress of the patterned metal sheet that is bonded to the ceramic. This obviously has a high influence on the cycling lifetime of the respective substrates. The thinning of silicon nitride substrate down to half that of the aluminium nitride (AlN) material, which allows an important reduction in thermal resistance to rival that of the thicker AlN, shows 45 % higher peak stress than for the AlN. This increase probably contributes significantly to the almost order of magnitude lower passive thermal lifetimes for the cycled Si_3N_4 substrates as shown in table-1. Even though the stresses will be lower for the AlN and

Al_2O_3 DBC substrates compared to the Si_3N_4 substrates, the cycling lifetimes are almost another order of magnitude lower. This indicates that the much higher flexural strength and fracture toughness of Si_3N_4, approximately twice as high as that of the other ceramics, is a major contributing factor to their much higher cycling lifetimes. The next step is emulate the active thermal cycling situation by modelling structures heated on the top surface by silicon carbide semiconductor devices and mounted and cooled on the bottom surface by a rigid cooler, in order to understand how the stresses compare to the passive cycling situation presented in this paper and how, from the point of view of the stress levels, passive cycles resemble real application conditions.

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