

Anomalous Reliability Behaviour of 99.99% and 99.999% Pure Aluminium Wire Bonds under Thermal Cycling

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Abstract

In this paper, we report on the reliability of ultrasonically bonded 99.99% (4Ns) and 99.999% (5Ns) pure aluminium wires under different passive thermal cycling ranges, namely -40°C to 190°C, -60°C to 170°C, -35°C to 145°C and -55°C to 125°C. In contrast to commonly accepted empirical relationships, the rate of bond strength degradation during cycling (determined through shear testing) was found to be more rapid in the wire bonds subjected to lower peak temperature (T_{jmax}) and lower temperature ranges (ΔT) for both wire types. In addition to these unexpected findings, the 4Ns wire bonds were found to be weaker than the 5Ns bonds under the cycling ranges investigated. In this paper, we give details of microstructural characterisation and nanoindentation hardness experiments undertaken to shed more light on these results. During thermal cycling, the development of grain size and morphology of the wires was found to be different for the various cycling ranges, and in particular they exhibited different levels of recrystallisation at the bond interface. These microstructural differences were corroborated by the hardness results and have been attributed to annealing effects occurring in the wire during the high temperature phase of cycling which remove some of the damage accumulated during the low temperature phase. A model has been proposed for the prediction of wear-out rate which incorporates both damage accumulation and damage removal mechanisms. We conclude that annealing effects which occur during thermal cycling can reduce damage accumulation and can influence wire bond reliability.

Key words: reliability, wire bonding, thermal cycling, microstructure, hardness, life prediction.

1.0 Introduction

During operation, power module interconnects are subjected to high temperature exposure and thermomechanical cycling. Under these conditions, the reliability of the power device is constrained by fatigue and creep processes which bring about the wear-out of the wire bonds and the degradation of the solder joints. Wire bonds are particularly susceptible to cracking and lifting off as a result of build up of plastic strain during cycling.

The characterisation of wear-out phenomena and the prediction of the failure of aluminium wire bonds during power device operation have traditionally been based on thermomechanical fatigue models. This is because the dominant failure cause is believed to be the substantial mismatch in thermal expansion

coefficient between the aluminium wires and silicon device onto which they are bonded (23.8×10^{-6} ppm K^{-1} and 2.6×10^{-6} ppm K^{-1} respectively), which creates thermomechanical strains at the bond interface during thermal cycling [1]. In order to assess wire bond wear-out, bonds are typically subjected to a constant thermal cycling range which can be described by a ΔT_j value (analogous to plastic strain amplitude) and a T_{mean} value. The mean number of cycles to failure N_{CM} , as given by the Coffin-Manson relationship, is strongly dependent on ΔT_j , as the following expression shows.

$$N_{CM} = A(\Delta T_j)^{-M} \quad \{1\}$$

where A and M are numerical constants usually determined empirically.

In general, workers have found this relationship to accurately predict the number of cycles to failure for scenarios where the peak temperature $T_{\max}$ during thermal cycling does not exceed 120°C [2, 3].

In this paper, however, we present findings which suggest that these thermomechanical fatigue models may not adequately describe the wear-out of wire bonds in cases where $T_{j\max}$ exceeds 120°C. Under these conditions, the degradation rate was found to be strongly influenced by $T_{j\max}$.

2.0 Experimental Details

99.999% (5Ns) and 99.99% (4Ns) aluminium wires, 375µm in diameter, were ultrasonically bonded to silicon dies coated with a 5µm thick aluminium top metal. The wires were bonded at room temperature. Bonded wires were then subjected to four different passive thermal cycling ranges, namely -40°C to 190°C, -60°C to 170°C, -35°C to 145°C and -55°C to 125°C. The degradation behaviour of the wire bonds was evaluated by measuring the bond shear strength at regular intervals. In addition, the development of microstructure, morphology and crack length of wire bond cross sections were characterised at regular intervals. Grain structure was revealed using FIB ion channelling in an FEI Quanta 3D CryoFIB/SEM. Nanoindentation of the wire bond cross-sections after 500 cycles was carried out using a MicroMaterials Ltd Nano test platform. A three-sided pyramid Berkovich indenter was utilised. Measurements were performed using an indentation load of 10 mN and dwell time of 30 s at maximum load. Semi-quantitative analysis of microstructures was performed using ImageJ software.

3.0 Results and Discussion

3.1 Crack Initiation in Bonded Wires

As reported elsewhere [4], prior to any thermomechanical exposure, the as-bonded wires often contain pre-existing heel cracks. These occur when the wires flex during loop formation during bonding, and usually open up at the interface between the wire and metallisation. Once thermal cycling begins, the cracks propagate, either at the same wire-metallisation interface, or in an intergranular fashion in the fine grained region until the wires lift off [5]. This increase in crack length is usually accompanied by a reduction in bond shear force, due to reduction in the bonded area. The micrographs in Figure 1 show increasing

crack length with increase in number of cycles in the -55 to +125°C range. The rate of crack propagation determines the failure life of the wire bond and can depend on a number of factors including bonding temperature [4, 5, 6].

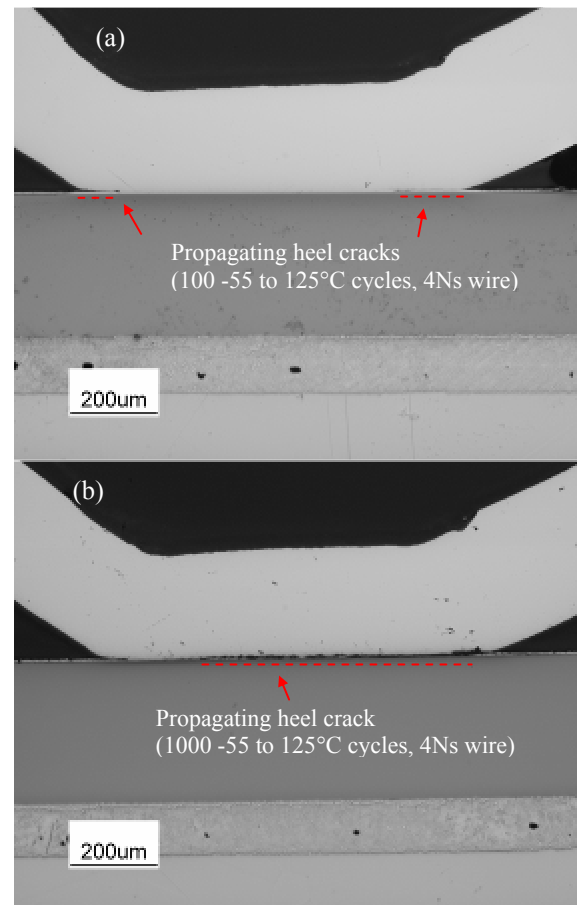


Figure 1. Optical micrographs of 4Ns wire bonds showing crack length after (a) 100 and (b) 1000 cycles.

3.2 Crack Propagation and Bond Shear Strength Degradation

Bond shear strength measurements taken at regular intervals during thermal cycling showed that 4Ns wire bonds degraded more rapidly than the 5Ns wires (Figure 2(a) and (b)). It was also interesting to observe that for both 4Ns and 5Ns wires, bonds subjected to higher maximum temperatures performed better (i.e. showed slower wear-out rates) than those subjected to lower $T_{j\max}$ despite the former exposures having larger ΔT_j values (230°C).

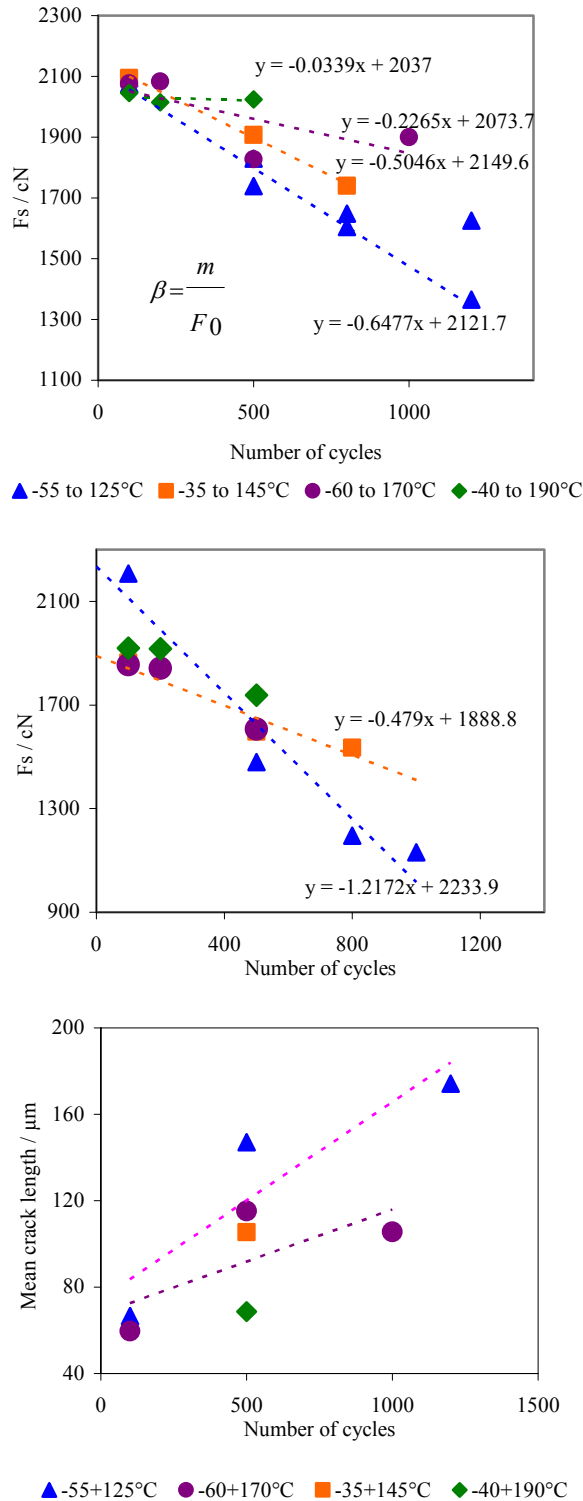


Figure 2 (a) & (b). Effect of increasing number of cycles on mean shear force for (a) 5Ns (b) 4Ns Al wire bonds. 2(c). Effect of increasing number on cycles on mean crack length of 5Ns Al wire bonds.

Crack propagation rate data from the same specimens showed a similar trend, with the fastest propagation rates occurring in the bonds cycled with the lowest ΔT (180°C) and lowest $T_{j\max}$ exposures (see Figure 2(c)). These experimental

results would appear to contradict conventional wire bond failure prediction models, such as the Coffin-Manson model [1, 7].

This seemingly anomalous observation has been attributed to ‘time at temperature’ or annealing effects. In other words, the slower bond degradation rate exhibited by the bonds exposed to higher ΔT and higher $T_{j\max}$ may be due to annealing processes which take place concurrently with damage accumulation during cycling. Typically, these annealing or ‘time at temperature’ effects would include thermally activated phenomena such as diffusion, dislocation annihilation, recrystallisation and grain growth, and would take place during the heating phase of a cycle. The longer heating excursion experienced by the higher $T_{j\max}$ sample is thought to favour/promote the progress of these annealing phenomena. These offset the damage which builds up during the cooling phase of thermal cycling (i.e. the build of plastic strain through dislocation multiplication). The overall rate of wear-out is thus believed to be the combined effect of these competing processes.

3.3 Microstructure, Morphology & Hardness

In order to elucidate the shear strength findings, the grain structures of 5Ns wire bonds subjected to two different thermal cycling ranges (namely -55 to 125°C and -60 to 170°C) were examined using FIB ion channelling, (see Figure 3 (a) to (f)). A number of observations were made. Firstly, the interfacial region in each wire bond was characterised by grain structure much finer than in the bulk wire. Secondly, in the ‘as-bonded’ condition, grains in the interfacial region were very fine and appeared to exhibit an elongated morphology (Figure 3 (a) and (d)). This is a sign of the initially deformed state in as-bonded wires. This deformed texture is expected to accelerate the rate of recrystallisation significantly during subsequent cycling [8, 9]. After 100 thermal cycles, the grain structure in both -55 to 125°C and -60 to 170°C wire bonds appeared more equiaxed (Figure 3 (b) and (e)). The grain structure of the specimen exposed to the lower peak temperature (125°C) appeared not to have coarsened noticeably; however, the grains in the specimen subjected to the higher peak temperature were evidently larger than in the as-bonded specimen.

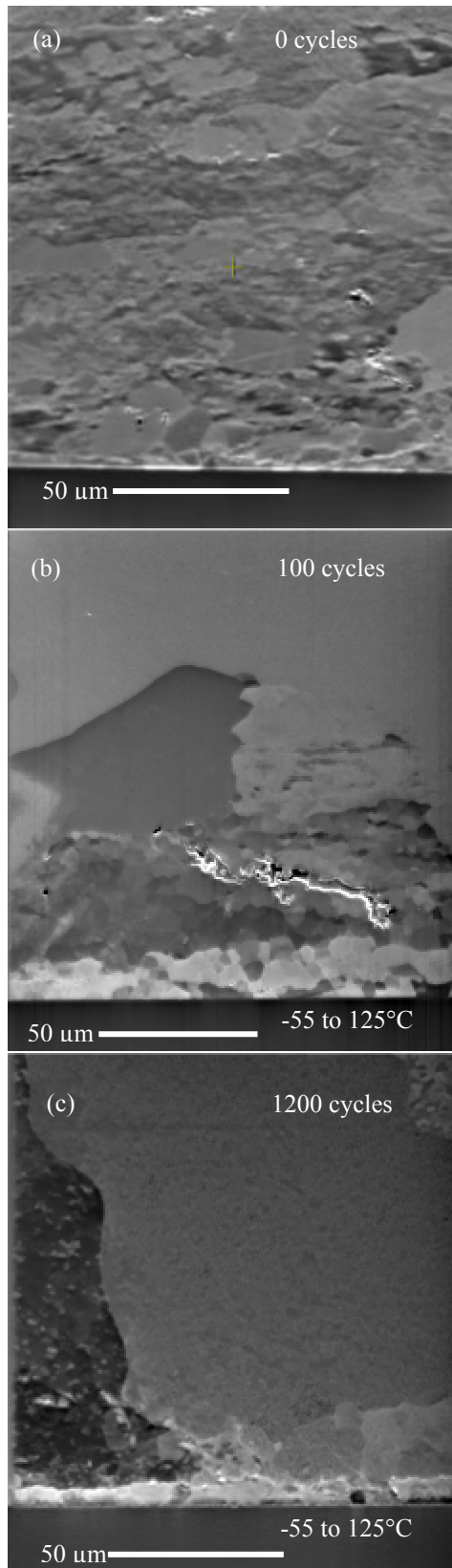


Figure 3(a) to (c). FIB images showing microstructures in interfacial regions after (a) 0 cycles, (b) 100 cycles and (c) 1200 cycles in the range -55°C to 125°C.

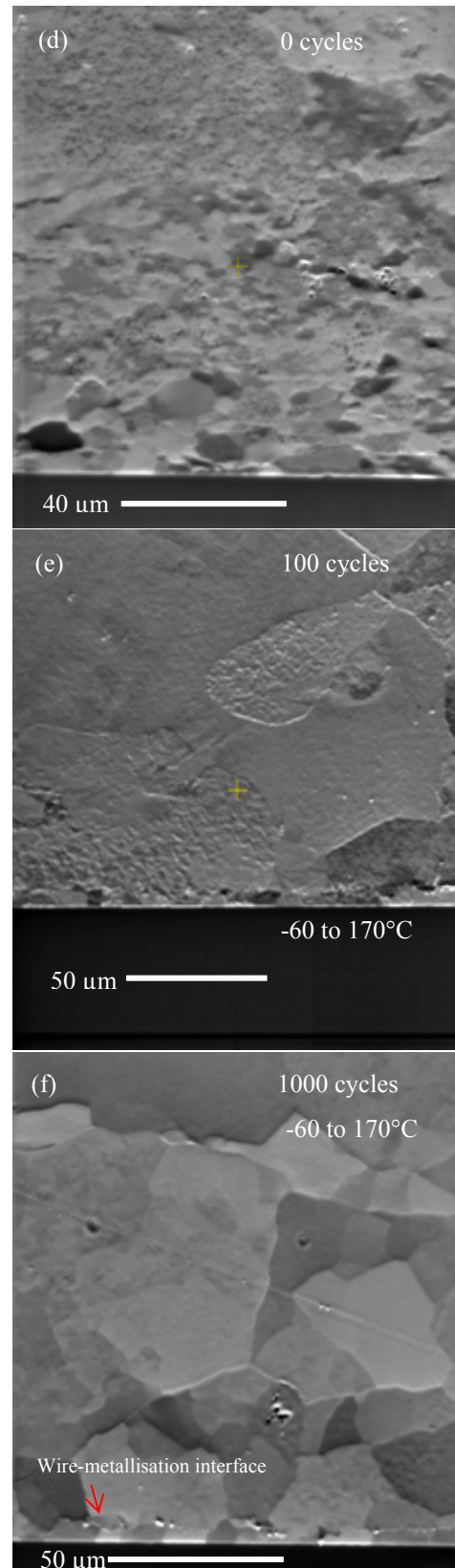


Figure 3(d) to (f). Microstructures of wire bond interfacial regions after (d) 0 cycles, (e) 100 cycles and (f) 1000 cycles in the range -60°C to 170°C.

After further cycling (1000 cycles), further grain growth was apparent in both cycling regimes. Moreover, overall coarsening near the bond interface appeared most pronounced for the bond subjected to the -60 to 170°C cycle. Although the lower T_{jmax} wire bond exhibited significant Ostwald ripening in the bulk wire, the grain structure near the bond interface remained fine in nature.

Some internal substructure was still visible after 1000 cycles in both wire bonds. This indicates that the wires were not fully recrystallised, in contrast, for example, to a 5Ns aluminium wire bond subjected to isothermal exposure at 350°C for 2 minutes (see Figure 4). Since recrystallisation and grain growth proceed relatively easily in these wires (given their high purity), the retention of partially recrystallised structure may indicate the continuing accrue ment of damage during cycling.

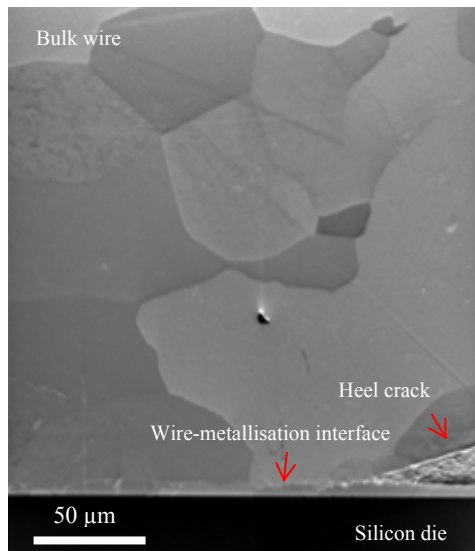


Figure 4. FIB image showing microstructure in interfacial regions of 5Ns Al wire bond subjected to isothermal exposure at 350°C for 2 minutes.

More pertinent to this discussion are nanoindentation hardness results of the interfacial region after 500 cycles, which revealed clearly different responses between the two thermal cycling regimes (see Figure 5). The bond subjected to the lower T_{max} (-55 to 125°C) was 20% harder (0.47-0.67 GPa) than the bond cycled from -60 to 170°C (0.41-0.50 GPa). This is indicative of a higher density of dislocations in the former and suggests more advanced recrystallisation and dislocation annihilation in the latter wire bond.

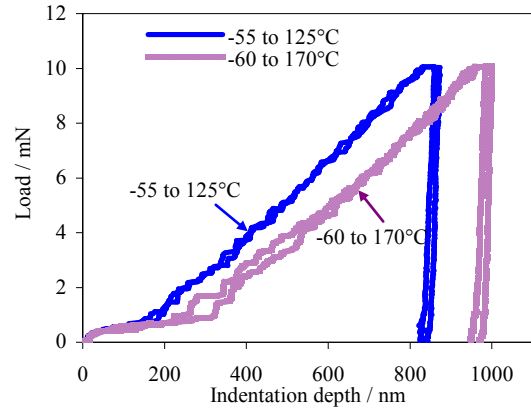


Figure 5. Nanoindentation of interfacial region of 5Ns Al wire bonds: load vs. displacement curves after 500 thermal cycles.

The microstructures and differences in hardness observed corroborate the suggestion that annealing or ‘damage removal’ processes occur to different extents in the two cycling regimes investigated, and are more pronounced at higher T_{jmax} values. The higher T_{jmax} cycle, which showed slower bond shear strength degradation rates, resulted in more softening than the lower T_{jmax} cycle. Furthermore, the bonds subjected to higher T_{jmax} values also exhibited larger interfacial grain size. These findings imply that the wire bond exposed to the higher T_{jmax} (170°C) has a lower density of dislocations and higher ductility, thus is better disposed both to reduce and absorb plastic strain during cycling than the harder structure of the lower T_{jmax} bond (125°C). This theory may also explain findings by previous workers that increasing the grain size within the interfacial region through annealing can slow down crack propagation [5, 6], as this evidently relates to reduction in dislocation density.

It also is possible that the higher T_{jmax} exposure promotes more interdiffusion and therefore a more robust metallurgical bond between the bulk wire and the aluminium metallisation on the silicon die. This would also have the effect of slowing down crack propagation.

3.4. Prediction of Wire Bond Wear-out Rate during Thermal Cycling

Typically, conventional life prediction models for wire bonds do not account for this damage removal which occurs in tandem with damage accumulation. This renders their application ineffective outside certain thermal cycling temperature limits.

To improve on the aforementioned empirical models which are based solely on the thermomechanical strains induced by ΔT , it is

proposed that the residual damage ΔD after n cycles within a wire bond is the result of the interaction between damage accumulated and damage removed during cycling. Damage accumulation is a function of plastic strain, dislocation density and minimum cycling temperature. Damage removal is thermally activated and depends strongly on $T_{\max}$.

$\Delta D = f(\text{damage accrued, fraction of damage annealed out})$.

{2}

During thermal cycling exposure, the development of bond shear strength can be expressed as

$$F_S = F_0(1 - \beta n)$$

{3}

where F_0 is the initial shear force, and β is a constant which describes the degradation rate of the shear force F_S after n cycles and can be obtained empirically for a given thermal cycling range (see Figure 2). The magnitude of β is therefore also determined by the relative amounts of damage annealed out and accumulated.

$\beta = f(\text{damage accrued, fraction annealed out})$

{4}

Assuming that the Coffin-Manson expression (equation {1}) accounts fully for the damage which builds up during cycling, and holds true for situations where the value of $T_{j\max}$ is sufficiently low such that fraction 'annealed out' is negligible, then β , can be approximated from N_{CM} in the following way:

$$\beta = \frac{1}{A\Delta T^{-M}}$$

{5}

For temperatures that are sufficiently high to cause significant damage removal the above expression can be modified to account for the damage that is annealed out during each cycle:

$$\beta = \frac{D_1}{A\Delta T^{-M}}$$

{6}

The term D_1 (< 1) gives the proportion of damage remaining after a single thermal cycle.

To provide an estimate of D_1 , consider a simple thermal cycle (as depicted in Figure 6), where τ is the time taken for a decline in temperature from $T_{\max}$ to $T_{\min}$. The proportion of damage that remains at the end of each cycle is determined by solving the following differential equation over one cycle:

$$\frac{dD}{dt} + Dk'_2 \exp\left(\frac{-Q}{RT}\right) = 0$$

{7}

In the above, Q and k'_2 are constants, the rate of decrease of damage being proportional to the product of damage remaining and a thermally activated diffusion parameter. Direct integration of {7} is complicated by the thermally activated term but for large values of Q , D_1 can be approximated as follows:

$$D_1 \approx \exp\left(-\frac{2k_2}{k_1}\right) \cdot \exp\left(\frac{2k_2}{k_1} \exp(-k_1\tau)\right)$$

{8}

The term k_1 describes the relationship between the amount of damage annealed out and $T_{\max}$ and k_2 is an annealing constant related to the rate of change of damage. k_1 and k_2 can be expressed as follows:

$$k_1 = \frac{Q\Delta T}{RT_{\max}^2 \tau}$$

{9}

and

$$k_2 = k'_2 \exp\left(\frac{-Q}{RT_{\max}}\right)$$

{10}

where the dominant damage removal process requires an activation energy Q .

There is good agreement between the fraction of damage remaining determined (i) numerically and (ii) from the approximations given in equation {8} for the different temperature ranges and reasonable activation energies, justifying use of the approximation. Furthermore, the wear-out rates predicted by these approximations of the anneal fraction show good agreement with experimental data when $Q = 84000 \text{ J mol}^{-1}$ (see Figures 7 and 8). It is interesting to note that this is similar to the activation energy for boundary diffusion in aluminium [9, 10].

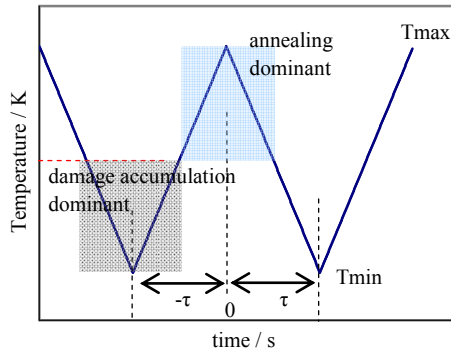


Figure 6. Schematic representation of a simple thermal cycle.

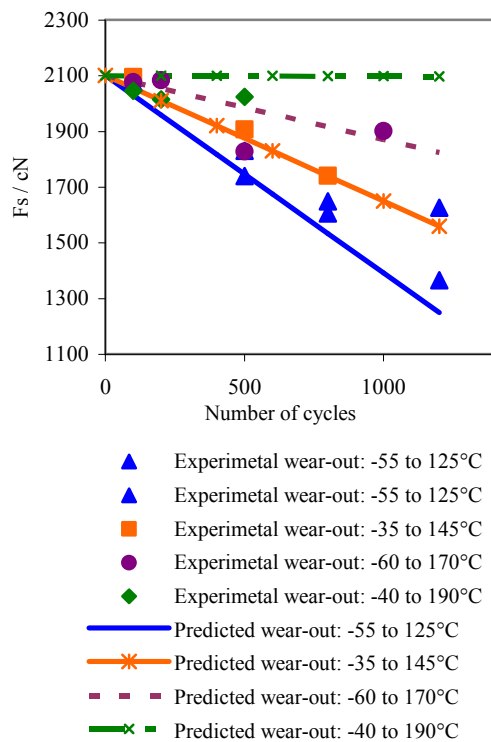


Figure 7. A comparison of predicted and experimental wear-out rates for 5Ns Al wire bonds.

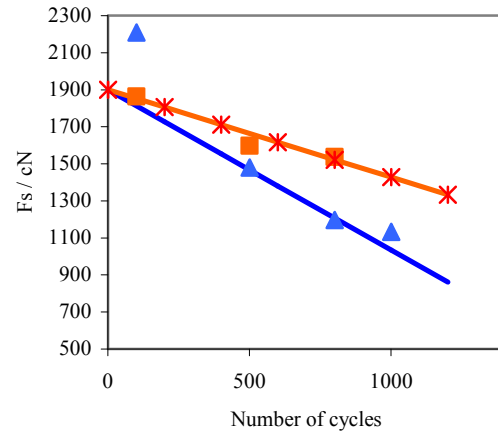


Figure 8. A comparison of predicted and experimental wear-out rates for 4Ns wire bonds.

4.0 Concluding Remarks

The microstructural response of pure aluminium wire bonds to thermal cycling is complex and is not adequately described by simple thermomechanical fatigue models. In addition to the build up of plastic strain ahead of the advancing crack, thermally activated phenomena also take place, and at sufficiently high peak cycling temperatures, these remove some of the damage accumulated. Microstructural investigations and nano-indentation data suggest that the extent of damage accumulation and removal during cycling depends on the thermal cycling range and the value of T_{jmax} in particular. The rate of change of residual damage depends on the combined effect of this damage build up and removal. A model has been proposed to predict wear-out rate which includes both damage accumulation and thermally activated damage removal mechanisms. There is very good agreement between predicted and empirically determined wear-out rates for both wire types and over the temperature ranges investigated.

These findings would suggest that aluminium wire bonds may be suitable for higher operation temperatures, due to these 'time at temperature' effects. However, further work is needed to characterise the damage accumulation and annealing mechanisms, in order to develop physics of failure models, from which wire bond life can be more accurately predicted.

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References

- [1] S. Ramminger, P. Türkes, G. Wachutka, "Crack mechanism in wire bonding joints", *Microelectronics Reliability*, vol. 38, nos. 6-8, pp. 1301-1305, 1998.
- [2] R. Fitzsimmons, E.H. Chia., R. Co and M.A. Sudbury, "Propagation mechanism and metallurgical characterization of 1st bond brittle heel cracks in AlSi wire", *IEEE Trans. on Components, Packaging, and Manufacturing Technology*, vol. 15, pp. 1081-1085, 1992.
- [3] A.T. Bryant, P.A. Mawby, P.R. Palmer, E. Santi, and J.L. Hudgins, "Exploration of Power Device Reliability Using Compact Device Models and Fast Electrothermal Simulation", *IEEE Transactions On Industry Applications*, vol. 44, No. 3, pp.894-903, 2008.
- [4] W. -S. Loh, M. Corfield, S. C. Hogg, and C. M. Johnson, "Thermal cycling reliability of heavy-gauge aluminium wires ultrasonically bonded at elevated temperature", *IMAPS 39th International Symposium on Microelectronics*, San Diego, USA, 2006.
- [5] W.-S. Loh, S. C. Hogg, R. J. Ikujiya, M. R. Corfield, P. Agyakwa and C. M. Johnson, "Effect of bonding temperature, post-bond heat treatment and extended range thermal cycling on the reliability of al wire bonds", *International Conference on High Temperature Electronics (HiTEC)*, Albuquerque, new Mexico USA, 2008.
- [6] J. Onuki, M. Koizumi and M. Suwa, "Reliability of thick Al wire bonds in IGBT modules for traction motor drives", *Advanced Packaging*, *IEEE Transactions*, vol. 23, no. 1, pp. 108-112, 2000.
- [7] S. Ramminger, N. Seliger and G. Wachutka, "Reliability Model for Al Wire Bonds subjected to Heel Crack Failures", *Microelectronics Reliability*, vol. 40, pp. 1521-1525, 2000.
- [8] R.D. Doherty, D.A. Hughes, F.J. Humphreys, J.J. Jonas, D. Juul Jensen, M.E. Kassner W.E. King, T.R. McNelley, H.J. McQueen, A.D. Rollett, "Current issues in recrystallization: a review", *Materials Science and Engineering*, vol. A238, pp. 219-274, 1997.
- [9] F.J. Humphreys, "Recrystallization and Related Annealing Phenomena" Elsevier Science Ltd, Oxford, UK, 1996.
- [10] N.Q. Chinh, P. Szommera, T. Csanadi, T.G. Langdon, "Flow processes at low temperatures in ultrafine-grained aluminium", *Materials Science and Engineering*, vol. A 434, pp. 326–334, 2006.