Correlation between mechanical properties and microstructure of different aluminum wire qualities after ultrasonic bonding

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1. Introduction

Ultrasonic bonding with heavy aluminium wires is well established in power electronic modules. To ensure a reliable functionality in application a variety of reliability studies have been realized [1–4]. As a result, the wire lift-off caused by a mismatch in the coefficients of thermal expansion (CTE) between the wire and substrate material appeared as one of the main failure mechanisms. In this case, the cyclic occurring thermo-mechanical loads lead to crack propagation in the wire material slightly above the metallization layer. As another failure mechanism, the wire heel-crack caused by mechanical deformation of the wire has been exposed [5,6]. These types of failure are influenced by the local mechanical properties of the bonded wire. During ultrasonic bonding, the wire material gets significantly changed by the deformation and ultrasonic induced hardening as well as softening mechanisms [7,8]. Due to the microstructural changes, the mechanical properties of the wire are modified.

To investigate the main failure mechanisms and to make an exact prediction of wire bond lifetime, it is crucial to analyse the microstructural evolution during ultrasonic wire bonding and to determine the resulting mechanical properties. The mechanical properties are of great interest since they can easily be used as input parameters for finite element simulation tools. Commonly, the mechanical properties of wire bonds are specified by the Young’s modulus and stress–strain data determined by tensile tests of the wire material [9–11]. Obviously, this approach does not consider the influence of the bonding process. First efforts to determine the mechanical properties of the bonded state by using nanoindentation are for instance presented in [12]. However, there is still a considerable ambiguity regarding a reliable determination of the mechanical properties and their dependency from the ultrasonic bonding process. In this respect, this article deals with the microstructural evolution during ultrasonic bonding and the resulting local mechanical properties of the wire.

2. Experimental procedure

2.1. Wire bonding

In this study, MOSFET chips wired with three different Al wire qualities are investigated. Fig. 1 shows the used MOSFET chip and a metallographically prepared longitudinal-section of a bonded wire. The nanoindentation test areas are schematically illustrated as blue frames. All wires were bonded with an Orthodyne wedge bonder on a 5 µm AlSi1Cu0.5 metallization layer deposited on a silicon chip which was attached using silver sintering. The investigated commercially available wires and their breaking loads are listed in Table 1.

The bonding parameters were adjusted with the aid of shear tests by generating shear forces and residues as high as possible without fracturing the chip.

2.2. Microstructural and mechanical tests

2.2.1. Electron backscatter diffraction (EBSD)

Microstructural investigations were carried out at metallographically prepared samples by EBSD. In a first step, the wire bonds were...
embedded near room temperature. After hardening at room temperature, the samples were carefully cut, mechanically ground and polished. As a final preparation step, all samples were polished and etched by ion beam. The EBSD measurements were conducted at an acceleration voltage of 20 kV and step sizes between 0.03 and 1 μm using a Digiview 4 CCD camera (EDAX) and SEM–Quanta FEG 400 (FEI).

2.2.2. Tensile tests and nanoindentation

To correlate the hardness with stress, tensile tests were performed up to several defined strains. The as-received and pre-strained wires were prepared as described in Section 2.2.1.

To analyse the local mechanical properties, load controlled nanoindentation tests were performed at the prepared longitudinal sections of the bonded wires. Furthermore, the as-received and pre-strained wires were examined. The used test system is a Triboindenter (Hysitron) with a Berkovich-tip. Each test was performed under the same test conditions. The maximum force was set to 9 mN and the duration of loading and unloading to 9 s, respectively. Before unloading, the maximum force was kept constant for 2 s.

3. Results and discussion

3.1. Microstructure of the as-received wires

The as-received wires were analysed regarding their crystallographic texture. As can be seen in Fig. 2, all investigated wires exhibit distinct but different textures.

These textures are the result of the specific wire drawing processes. In the case of Q1, the majority of grains exhibit a (001)–A1 orientation. The Q2 wire exhibits a (001)–A1 and additionally weak (111)–A1 texture component. In contrast to Q1 and Q2, the pole figures of the Q3 wire show a clear (111)–A1 fibre texture. Furthermore, the Q3 wire shows a weak gradient in grain orientation regarding the wire diameter. Slightly below the wire surface, the (111)–A1 texture intensity seems to decrease.

3.2. Microstructure after wire bonding

The microstructural change of the different wire qualities after ultrasonic bonding is shown in Fig. 3.

After wire bonding a distinct gradient in grain orientation and size has evolved within the wedge area. Remarkably, each wire bond exhibits a distinct (101)–A1 textured area slightly above the metallization. This texture is considered a rotated cube (RC) texture. The black frame within the inverse pole figure maps shows the areas used for the measurements of the RC texture. Within the investigated longitudinal sections, Q1 seems to exhibit the highest magnitude of the RC textured area. Between Q2 and Q3 no obvious difference regarding the magnitude of RC textured area can be observed. Above the RC textured area and especially slightly below the wire upside the as-received wire texture is still extant. Considering the heel area, no remarkable change in texture can be observed. These characteristic microstructural gradients occur nearly in the same manner for Q1, Q2 and Q3.

The evolution of RC textures is already discussed in [13] as a result of microstructural processes such as dynamic recrystallization, recovery and plastic deformation during ultrasonic bonding. Since RC textures can be observed for different Al wire qualities, it can be assumed that their evolution is generally possible for face centered cubic (fcc) metals. Beside the assumed transferability to other fcc metals, the occurring of RC texture might have considerable influence on the results of shear tests which were commonly used for the quality control. In conclusion, in addition to geometrical aspects, the local material properties, especially the RC texture, should be considered in the interpretation of shear results.

3.3. Mechanical characterization

3.3.1. Determination of elastic properties

Each as-received wire exhibits a fibre texture. However, wire bonding leads to a texture gradient in the wedge area. This texture gradient consists of RC texture slightly above the metallization and fibre texture at the upside of the wedge. The intensity of both depends on the wire bonding process. In this case, all wire bonds exhibit a distinct RC texture and minor changed fibre texture. Also the texture of the heel area appears to be equal to the as-received wires. Consequently, the texture of the heel area and area at the upside of the wedge can be assumed as equal to the texture of the as-received wires. In the following part, we correlate the observed fibre and RC textures with the respective elastic properties.

Fibre textures lead to transversely isotropic behaviour. Using the single crystal compliance tensor $S_{ijkl}$ of Al and the texture information from Fig. 2, the elastic properties can be approximated [14]. In the case of transverse isotropy, the elastic properties are represented by the Young’s moduli parallel and perpendicular to the A1 axis ($E_{A1}, E_{A2}, E_{A3}$).
Furthermore, no clear texture perpendicular to A1 can be observed. Therefore, Q2 is assumed as isotropic on a macroscopic level.

Complex mathematical expressions for the calculation of the shear plane and direction dependent shear modulus for cubic crystals are given in [17,18]. However, for simplification the general expression of Hearmon [16] seems to be more convenient. In this expression the shear modulus of a specific plane is related to torsion around all directions within this plane. Using this expression (see Eq. (2)), the same approach as used for the approximation of the representative Young’s moduli was applied to approximate the representative shear moduli.

The tension plane and direction dependent Poisson ratio was calculated with Eq. (3) [17]. Applying this equation, we receive the respective Poisson ratios \( \nu_{[100][010]} = 0.36, \nu_{[10\overline{1}][110]} = 0.34, \nu_{[001][110]} = 0.41 \) and \( \nu_{[\overline{1}0\overline{1}][110]} = 0.27 \). As a consequence, the Poisson ratios referring to the in-plane contraction for the \( [010] \) and \( [111] \) crystal directions which were aligned parallel to the tension direction are set constant to 0.36 and 0.34. Although, \( \nu_{[h'k'\ell'][[110]} \) varies corresponding to the respective in-plane direction within the crystal system, we assume that \( \nu_{[h'k'\ell'][[110]} \) is nearly constant within the sample system. This assumption is made by considering the random orientation of the in-plane directions within the sample system (see Fig. 2). Here, we note that this assumption should also be valid for \( \nu_{[h'k'\ell'][[010]} \) and \( \nu_{[h'k'\ell'][[110]} \). Averaging the minimal and maximal Poisson ratios for the in-plane directions of the \( (110) \) plane results in 0.34. Using this approach, it can be seen that the calculated Poisson ratios referring to the \( (010) \), \( (111) \) and \( (110) \) planes vary between 0.34 and 0.36. Under consideration of the corresponding area fractions the difference between the wire qualities is minor. The results for the approximation of the elastic properties are summarized in Table 2.

In the case of Q1, the maximum difference between the sample direction dependent Young’s modulus is \(-4\) GPa which is \(-6\%\) of the...
maximum value. Comparing Q1 and Q3, the maximum difference occurs in the A1 direction with ~8 GPa which is ~11%. Despite of the relatively low anisotropy of the Al single crystal, the calculated percentage differences appear quite high. Therefore, considering the elastic anisotropy and the difference between wire qualities should improve the significance of finite element simulations for wire bonds under various loading conditions. Especially in the case of loading conditions with a major elastic part, the consideration of the current results might have considerable impact. Also, it should be mentioned that in the case of more anisotropic materials such as copper, the direction dependency would be more significant by far.

Due to the high intensity of RC texture, the area slightly above the metallization layer probably exhibits anisotropic behaviour. Simplifying, the elastic properties of this area can be described by a tilted single crystal. Under this assumption, the elastic properties can be given as a 6 × 6 stiffness tensor which can be directly implemented into finite element simulation tools. However, it is necessary to tilt the stiffness tensor according to the respective texture. The RC texture corresponds to a 45° tilt around the A3 axis. Applying the rules for transformation of the stiffness tensor, the 45°/A3 tilted stiffness tensor can be determined. This tensor transformation results in

\[
S_{11} = S_{22} = 113.3, \quad S_{12} = S_{21} = 56.7, \quad S_{13} = S_{31} = S_{23} = S_{32} = 62, \quad S_{33} = 108, \quad S_{44} = S_{55} = 28.3 \quad \text{and} \quad S_{66} = 23.\]

3.3.2. Correlation between hardness and microstructure

Fig. 4 summarizes the results for the nanoindentation tests of the as-received and bonded wires. The diagrams show the wire hardness measured in the wedge area in regard to the distance from the Si interface. Furthermore, the hardness of the as-received wires and loop area is indicated as a constant value (horizontal lines). For all wire qualities, the lowest hardness within the wedge area is located slightly above the Si interface in the area of RC texture. This indicates the occurring of

| Table 2
<table>
<thead>
<tr>
<th>Elastic properties</th>
<th>Q1</th>
<th>Q2</th>
<th>Q3</th>
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<tr>
<td>$E_{A1}$</td>
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<td>69</td>
<td>70</td>
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<tr>
<td>$E_{A2}$</td>
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<td>$\nu_{A1A3}$</td>
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* Isotropic properties.

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dynamic recrystallization and recovery during the ultrasonic deformation. Except for Q3, the wedge clearly exhibits higher hardness than the loop area. In all cases, the loop area exhibits slightly higher hardness than the as-received state. Surprisingly, in the case of Q3 the hardness of RC textured area is nearly the same as of the loop area and the as-received state.

3.3.3. Correlation between hardness and stress

The hardness values measured at the as-received wires and defined interrupted tensile test samples were correlated to tensile stresses on the basis of the Tabor correlation [19]. In this correlation, each hardness value corresponds to a defined strain level which is the sum of the initial plastic strain $\varepsilon_{pl}$ adjusted by the tensile test and the representative strain from indentation $\varepsilon_{r}$. The representative strain depends among others on the indenter tip geometry and the mechanical properties of the sample material. For simplification we set the representative strain to constant 0.02 which is within the range of the calculated values by [20] and [21] for a sharp indenter and a metallic sample material. It should be noted that this approximation leads to an uncertainty regarding the absolute stress value.

Due to the indentation size effect (ISE), the hardness and therewith the constraint factor depend on the indentation depth [22]. Consequently, the calculated constraint factor is only valid for hardness values determined under the same test conditions as described in Section 2.2.2. The correlation between the experimental stress–strain curves and the hardness values results in a constraint factor of around $c = 6.7$–8.1 for Q1, $c = 4.2$–5.4 for Q2 and $c = 4.2$–5 for Q3 (see Fig. 5). Applying the respective constraint factor, it is possible to calculate the local representative stresses for each measured location within the wedge and loop area. Due to the relatively small representative strain of 2%, the calculable representative stresses can be approximated as flow stresses for the initiation of plastic deformation. As an example, for the RC textured area of Q1 we calculate flow stresses between 51–62 MPa which is twice as much as the flow stress of the as-received wire. These results show the necessity of considering the change of material properties during the ultrasonic bonding process.

4. Conclusions

Despite of the different initial texture, grain size and hardness, all investigated wires exhibit a rotated cube textured area after wire bonding. Hardness measurements show that the bonding process increases the hardness of the wedge and loop area. Due to dynamic recrystallization and recovery, the RC textured area exhibits the lowest hardness within wedge for all wire qualities. The calculated local elastic and plastic material properties indicate the importance of considering the bonding process and crystallographic wire texture for accurate lifetime predictions of wire bonds in power modules. Furthermore, the results form a basis to improve existing simulation models.

Acknowledgements

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References


