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Diffusion Welding of Copper to Aluminum †

Toshio ENJO*, Kenji IKEUCHI and Naofumi AKIKAWA*****

Abstract

Diffusion welding of copper to aluminum has been carried out in a vacuum environment. The microstructure of the bonding zone is examined in detail with several metallographic methods to make clear the important factor which affects the mechanical properties of the joint. Results obtained are summarized as follows :

1) An intermetallic compound layer is observed which can be divided into three characteristic region I, II and III. The intermetallic compounds of θ and γ_2 are formed in I and II region, respectively. The hardness of region II is the highest of the regions.

2) The growth of the intermetallic compound layer is considered to be controlled by the atomic diffusion and the increase in the intimate contact area between the faying surfaces. The diffusion process becomes more important with the increase in welding temperature.

3) In the earlier stage of welding (shorter welding time), the tensile strength of joint increases with the rise of welding temperature, time and pressure. This stage is considered to be a process where the intimate contact between the faying surfaces is developed.

4) In the later stage of welding (longer welding time), the tensile strength of joint approaches to a constant value (2 ~ 3 kg/mm²) much lower than that of aluminum base metal. This tendency is observed for the thickness of the intermetallic compound layer greater than 15 ~ 20 μm. In this case, fracture is developed in the intermetallic compound layer, but not at the welding interface. The strength of joint welded in this stage is considered to be controlled by the strength of the intermetallic compound layer.

KEY WORDS: (Diffusion Welding) (Dissimilar Materials) (Aluminium) (Copper) (Intermetallics)

1. Introduction

It is generally accepted that the diffusion welding is suited to welding between dissimilar metals which is particularly difficult by conventional techniques [1~4]. Many investigations have been reported about the diffusion welding between dissimilar metals [1~12]. However, with some dissimilar metal combinations, the strength of joints is much lower than that of the base metal because of the formation of brittle intermetallic compounds in the bonding zone.

The authours have already reported about important factors for the strength of joint between several dissimilar metal combinations which form intermetallic compounds [9~12]. In the present investigation, the diffusion welding of aluminum to copper has been carried out as an example of joint between dissimilar metal combination which forms intermetallic compounds. With several metallographic methods, the intermetallic compounds formed in the bonding zone have been identified and the growing kinetics of the intermetallic compound layer has been examined. The effects of the intermetallic compound layer on the joint strength have been discussed.

2. Experimental Details

The base metals used in this investigation were commercially pure copper and aluminum, whose chemical compositions are shown in **Table 1.** The specimen for the observation of microstructure has the shape of a cubic whose edge length is 3 ~ 4mm. The faying surface was polished with a 1500 grade emery paper and degreased by washing in acetone. The microstructure of the bonding zone was observed in the course of welding with a high temperature microscope (HTM) equiped with a compressing devise. The surface for observation with the HTM

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Table 1 Chemical compositions (wt%).

	Cu	Fe	Si	Mg	Mn	Zn	Cr	Ti	Al
Al	0.01	0.24	0.10	0.01	0.01	0.04	0.01	0.01	Bal.

	Pb	Fe	Ni	S	Ag	O_2	P	Cu
Cu	.0004	.0007	.0005	.0019	.0011	.0415	—	Bal.

was finished by polishing after preliminary welding for 5 or 120 min at the same welding temperature and pressure as those where the observation was to be carried out. A tungsten wire (20μm in diameter) was inserted between the faying surfaces as a marker prior to welding.

Joints to examine the tensile strength were prepared by welding cylindrical specimens. The diameter and length of the specimen were 20mm and 37mm, respectively. The specimens were welded with the end planes of the cylinders as faying surfaces. The faying surface was machined to 3-S in JIS number with a lathe and degreased by washing in acetone. The welding of the cylindrical specimen was carried out in a vacuum environment with the same apparatus as that reported in a previous paper[10]: the bonding zone was heated with a high frequency induction heater and the welding pressure was applied with a hydraulic press. Most of the joints were cut into the shape as shown in Fig. 1 for tensile test. As for

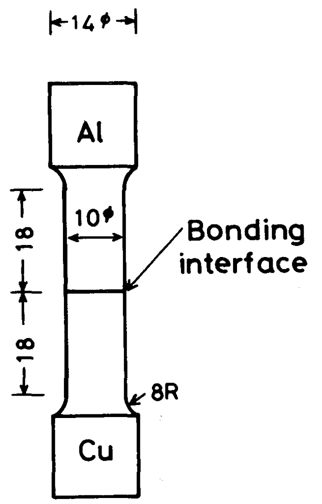


Fig. 1 Specimen for tensile test.

the joint with very low strength, the tensile test was carried out without shaping the joint into the tensile test specimen, since such joint was often fractured in the course of shaping. Both the joints for the observation of microstructure and for the tensile test were welded in a vacuum environment of $10^{-4} \sim 10^{-5}$mmHg.

3. Results and Discussion

3.1 Microstructure of the Bonding Zone

Photo. 1 shows the microstructures of the bonding

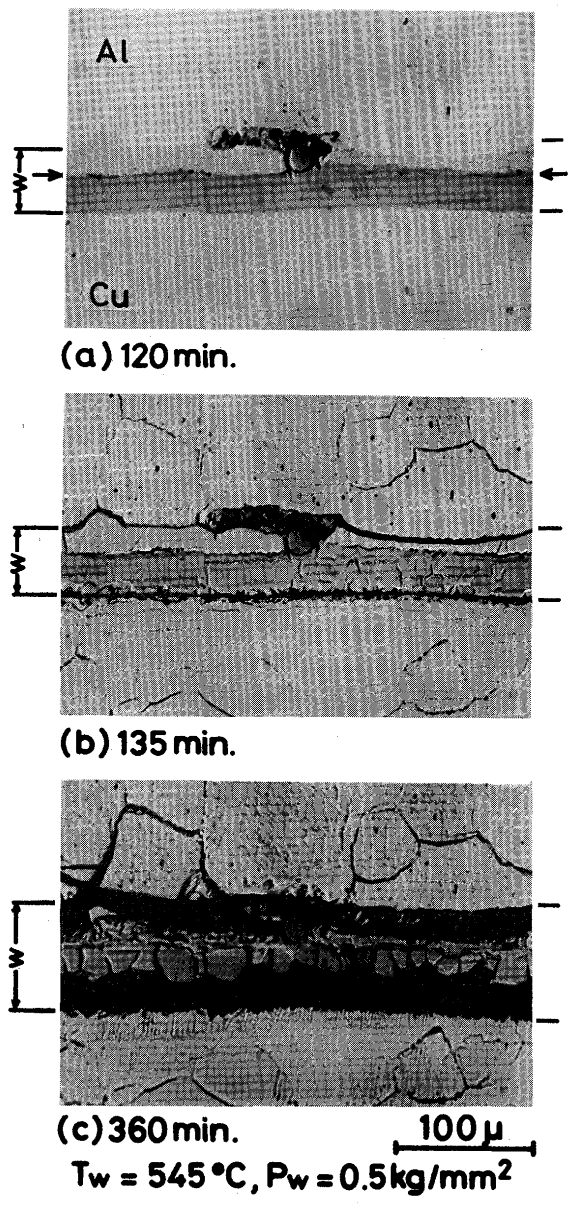


Photo. 1 Microstructures for a bonding zone between copper and aluminum observed with a high temperature microscope. T_W and P_W are welding temperature and pressure, respectively.

zone observed with the HTM. The microstructure after the preliminary welding for 120min at 545°C is shown in Photo. 1(a). As shown in Photo. 1(a), an intermetallic compound layer and the marker of tungsten wire were observed in the bonding zone. The arrows indicated the initial bonding interface. In the bonding zone after welding for 135min, grain boundaries revealed by thermal etching were observed both in aluminum and copper matrix, as shown in Photo. 1(b). In the bonding zone after welding for 360 min, the black bands were observed on both the side of the intermetallic compound layer, as shown in Photo. 1(c). These black bands are regarded as the regions where the intermetallic compound layer had grown after the preliminary welding.

Photo. 2(a) and (b) show the superficial microstructure

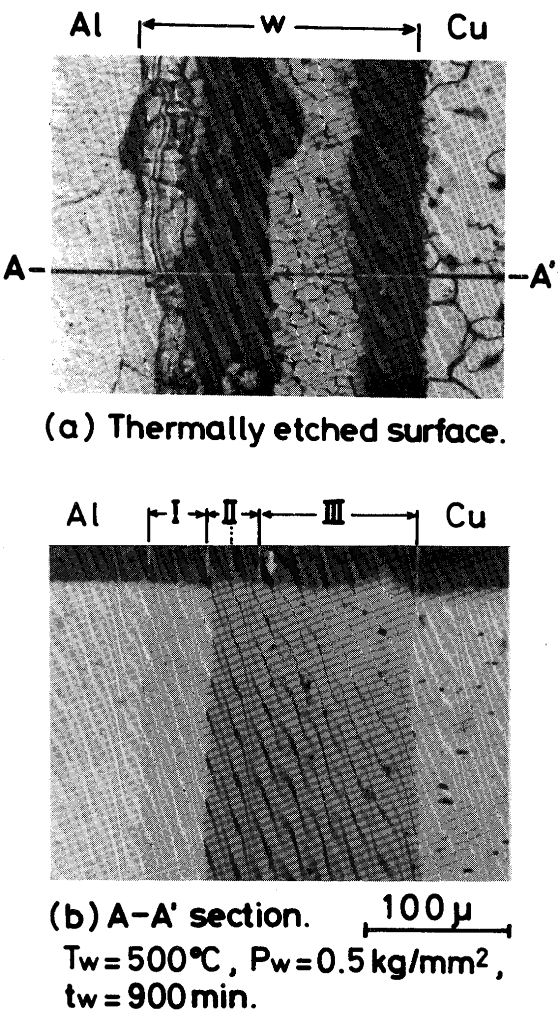


Photo. 2 Microstructures for a bonding zone between copper and aluminum. T_W, P_W and t_W are welding temperature, pressure and time, respectively.
(a) Superficial microstructure.
(b) Microstructure for the A-A' section in Photo. (a).

observed with the HTM and the internal microstructure sectioned along the line A-A' in (a), respectively. As shown in Photo. 2(a) and (b), the width of the intermetallic compound layer observed with the HTM is nearly equal to that of internal microstructure. The intermetallic compound layer can be divided into three characteristic region I, II and III, as shown in Photo. 2(b). Fig. 2 shows the distribution of Knoop's hardness numbers in the bonding zone. As shown in **Fig. 2**, the hardness of the intermetallic compound layer was much higher than that

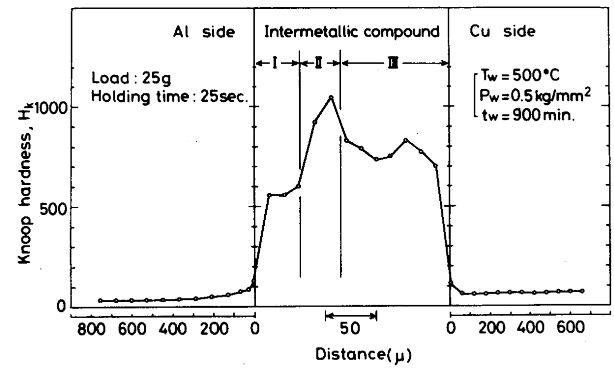


Fig. 2 Distribution of Knoop's hardness numbers in bonding zone.

of aluminum and copper matrix, and the region II was the hardest of the regions.

In order to identify the intermetallic compounds formed in each region, the bonding zone was examined with an electron probe X-ray microanalyser (EPMA) and X-ray diffraction analysis. **Fig. 3** shows the distributions

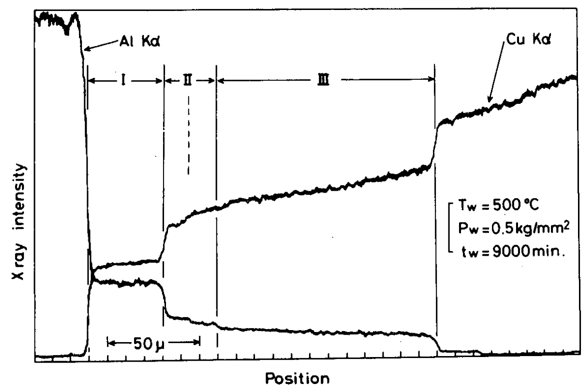


Fig. 3 Distribution of each element in bonding zone analysed with an EPMA.

of characteristic X-ray intensity of copper and aluminum analysed with the EPMA. The distribution curves of copper and aluminum in Fig. 3 show that the intermetallic compound layer can be divided into three characteristic region I, II and III in the same manner as the microstructure shown in Photo. 2(b). And they also suggest that the region II can be divided into two more regions at the position indicated by the broken line in Fig. 3.

Fig. 4 shows X-ray diffraction patterns obtained on the

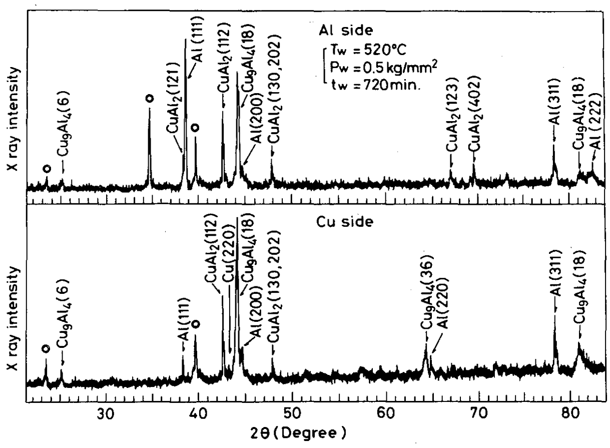


Fig. 4 X-ray diffraction patterns from the fractured surfaces of the joint between copper and aluminum fractured at the bonding zone.

fractured surfaces of a joint which was broken at the bonding zone. The X-ray analysis was carried out with Cu-K_α radiation. The diffraction lines of θ ($CuAl_2$) and γ_2 (Cu_9Al_4) were observed on both the fractured surfaces of copper side and aluminum. According to an equilibrium phase diagram[13] of the binary system between copper and aluminum, the five intermetallic compounds of θ ($CuAl_2$), η_2 (CuAl), ζ_2 (Cu_4Al_3), δ (Cu_3Al_2) and γ_2 (Cu_9Al_4) are formed in the welding temperature range of this investigation. Consequently, the region I and III are considered to consist of θ and γ_2 phase, respectively. Three diffraction lines (marked with◎) which could not be related with any phase are considered to correspond to any of the intermetallic compound of η_2, ζ_2 or δ phase. However, it is said that not all of the intermetallic compound formed in equilibrium system are formed in nonequilibrium systems such as diffusion couple[14].

In order to investigate the growing process of the intermetallic compound layer, the thickness of the intermetallic compound layer was observed with the HTM in the course of diffusion welding. For some dissimilar metal combinations, it is reported that the growing of the intermetallic compound layer causes the weakening of the mechanical property of joint[6, 7, 8]. **Fig. 5** shows the thickness of intermetallic compound layer (w) as a function of welding time (t). As shown in Fig. 5, the growing process can be described by the equation as

$$w = k \cdot t^m \tag{1}$$

where k and m are constant. It is said that the exponent m in eq. (1) equals to 0.5 when the growing rate of the intermetallic compound layer is controlled by the atomic diffusion[15]. As shown in Fig. 5, the exponent m was 0.78 ~ 0.58 and approached to 0.5 with the rise of

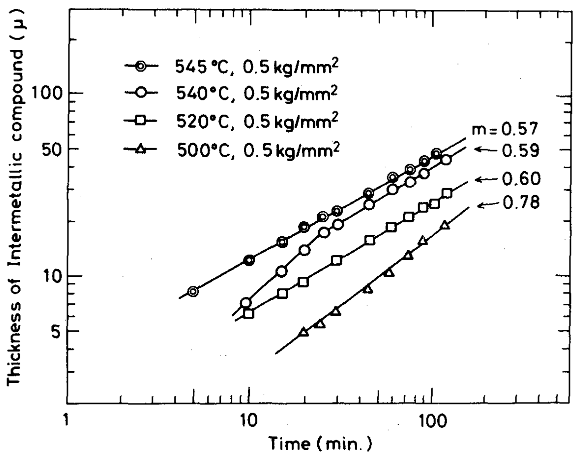


Fig. 5 The thickness of intermetallic compound layer as a function of welding time at various welding temperatures.

welding temperature. This indicates that the growing process of intermetallic compound layer was mainly controlled by the atomic diffusion at welding temperatures above 545°C, but the effect of any process other than atomic diffusion became more important as the welding temperature decreased. **Fig. 6** shows the effect of

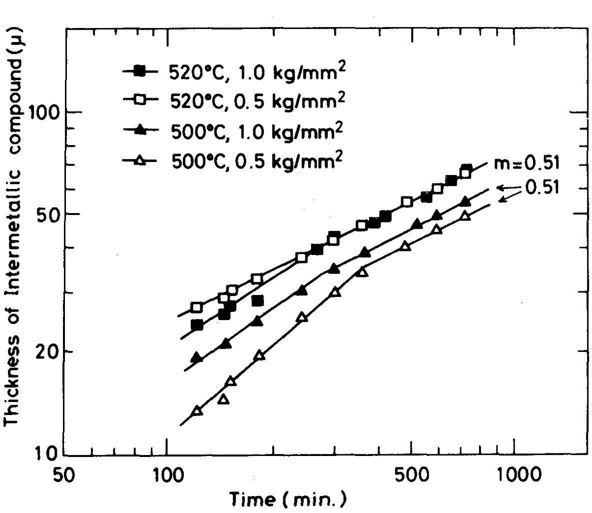


Fig. 6 Effect of welding pressure on the growing process of intermetallic compound layer at the welding temperature of 500° and 520°C.

welding pressure on the growing process. At the welding temperature of 500°C, the growing rate of the intermetallic compound layer was accelerated by increasing the welding pressure as shown in Fig. 6. On the other hand, at 520°C the effect of welding pressure on the growing rate was negligibly small. These results indicate that the attainment of intimate contact between the faying surfaces was an important process which affected the growing rate of intermetallic compound layer at lower welding temperatures.

3.2 Tensile Test

Fig. 7 shows the variation of the tensile strength of

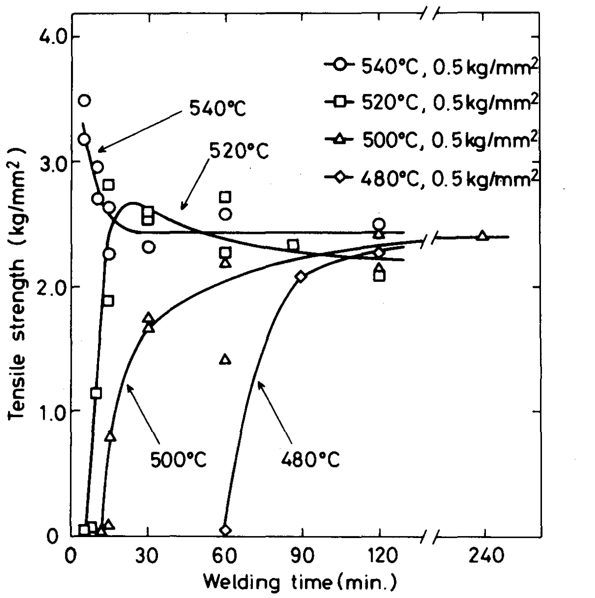


Fig. 7 Tensile strength of joint as a function of welding time at the welding pressure of 0.5 kg/mm^2 and various welding temperatures.

joint with welding time at various welding temperatures. The tensile test was carried out with Instron-type machine at the strain rate of 5.6×10^{-4} sec^{-1} at room temperature. As shown in Fig. 7, the tensile strength of joint increased more rapidly with the rise of welding temperature. However, for longer welding time (later stage), the tensile strength of joint welded at any temperature approached to the constant value of 2~ 3 kg/mm^2 which was much lower than that of aluminum base metal (6 ~ 8 kg/mm^2). When the joint had the nearly constant value, the thickness of the intermetallic compound layer was estimated to be greater than 15 ~ 20μm for each welding temperature from Fig. 5 and 7. This suggests that in the later stage the joint strength was controlled mainly by the strength of intermetallic compound layer. In order to confirm this suggestion, the fractured surfaces of joints welded in the later stage were observed with a scanning electron microscope (SEM). **Photo. 3(a) and (b)** show secondary electron images of the fractured surface of aluminum side observed with the SEM. The distribution of aluminum characteristic X-ray intensity is also shown in Photo. 3(a). It is considered that in region A fracture had occured along the bonding interface, because the traces of grooves caused by the finishing of the faying surface with a lathe were observed in region A as shown in Photo. 3(a). Photo. 3(b) shows the boundary between region A and B observed at a higher magnification. This photograph indicates that in region B fracture occured at a place which deviated a little to the aluminum base metal from the bonding interface. This is supported by the fact that the intensity of aluminum characteristic X-ray was lower in region A than in region B. A cross sectional microstructure for the fractured zone of a joint welded in the later stage is shown in **Photo. 4.** As shown in Photo. 4, the fracture occured at the intermetallic compound layer. From these results, it is considered that in the region B the fracture occured at the intermetallic compound layer but not at the bonding interface. The fraction of region B increased as the welding time increased and the tensile strength of joint approached to the constant value. Thus, the joint welded in the later stage fractured mainly at the intermetallic compound layer but not at the bonding interface. Consequently it is concluded that the tensile strength of joint between copper and aluminum was controlled by the strength of intermetallic compound layer in the later stage.

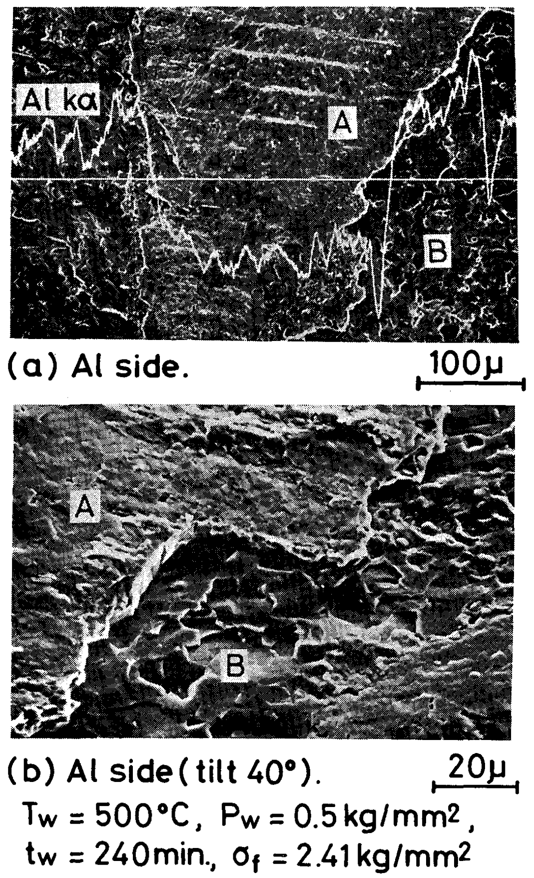


Photo. 3 Scanning electron micrographs for the fractured surface of aluminum side. The intensity of characteristic X-ray of aluminum analysed along the white straight line is shown in Photo. (a).

When a joint with a very thick intermetallic compound layer was cooled rapidly after welding, cracks were sometimes observed in the intermetallic compound layer

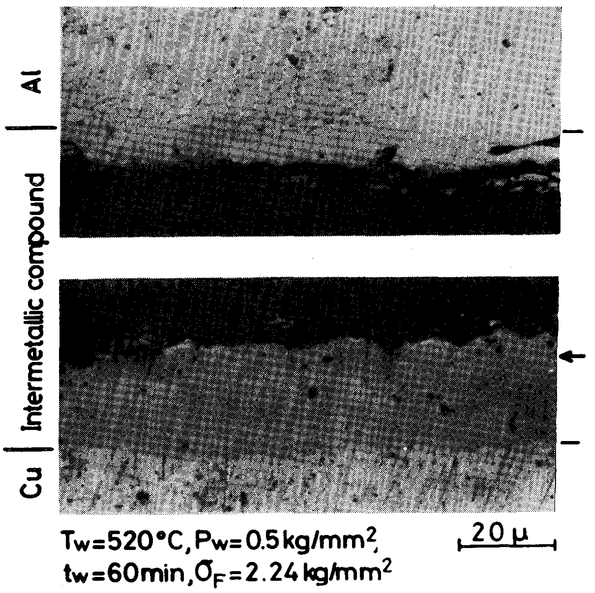


Photo. 4 Cross sectional microstructure for the fractured zone of copper and aluminum joint.

as shown in **Photo. 5**. As shown in Photo. 5, the cracks propagated along the region II which is the hardest region in the intermetallic compound layer (see Fig. 2). These

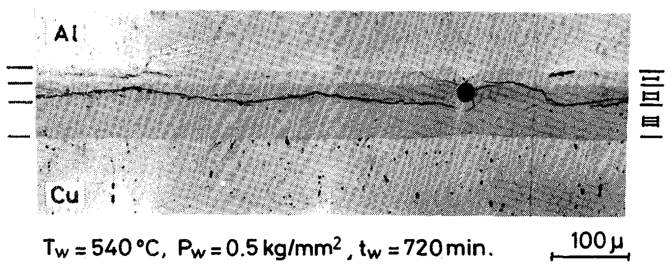


Photo. 5 Microstructure for the bonding zone between copper and aluminum with cracks in the intermetallic compound layer.

cracks are supposed to be due to thermal strain in cooling process which is caused by the difference of the coefficient of thermal expansion between the phases in the vicinity of bonding interface. Considerable care must be taken to cool slowly a joint with a thick intermetallic compound layer.

For shorter welding time (earlier stage), the tensile strength of joint increased with the rise of welding time as shown in Fig. 7. **Fig. 8** shows the effect of welding pressure on the tensile strength of joint welded in the earlier stage. The tensile strength of joint depended strongly on welding pressure and increased more rapidly by increasing the welding pressure. Such effect of welding pressure on the joint strength indicates that in the earlier

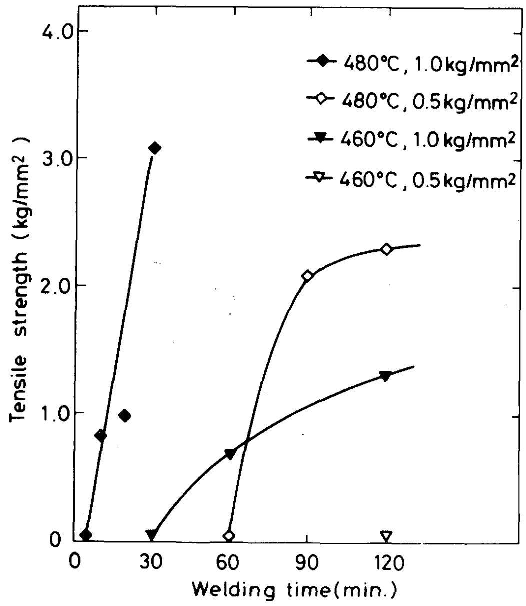


Fig. 8 Effect of welding pressure on the tensile strength of joint at the welding temperature of 460° and 480°C.

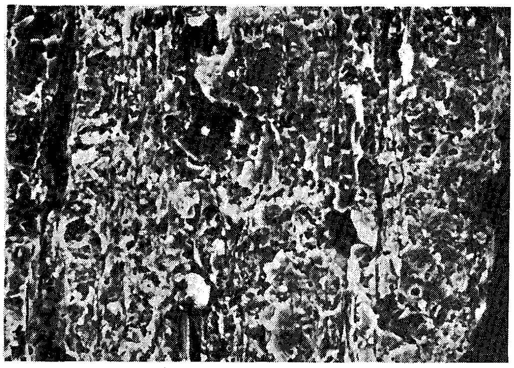
stage of diffusion welding between copper and aluminum the increase in joint strength was controlled by the attainment of intimate contact between the faying surfaces due to creep deformation in the bonding zone.

Photo. 6 shows a fractured surface of aluminum side for a joint welded in the earlier stage. The white circular regions were observed as shown in Photo. 6(a). These white regions are considered as the place where the bonding was achieved preferentially and to be formed along the microasperities caused by the finishing of faying surface with a lathe. This fact supports that the earlier stage of diffusion welding between copper and aluminum is the process where the intimate contact area between the faying surfaces increases.

As shown in Fig. 7, peaks were observed in the relation between joint strength and welding time at the welding temperature of 520° and 540°C. These peaks can be explained as follows: at welding temperatures above 520°C, the intimate contact between the faying surfaces was developed rapidly before the thickness of intermetallic cmpound layer grew largely and had harmful effect on the joint strength.

The authors have already pointed out the significance of the harmful effect of aluminum oxide film on bonding process in a previous paper[16]. The similar effect of aluminum oxide film is expected on the bonding process between aluminum and copper. That is, it is expected that

(a) Al side. 200μ



(b) Al side. 20μ
T_w = 520 °C, P_w = 0.5 kg/mm^2,
t_w = 15 min., σ_f = 1.78 kg/mm^2

Photo. 6 Scanning electron micrographs for the fractured surface of aluminum side.

the aluminum oxide film on the faying surface inhibits the real metallic contact between the faying surfaces. However, it is considered that the effect of the aluminum oxide film is confined to only the earlier stage and in the later stage the joint strength is controlled mainly by the strength of the intermetallic compound layer.

4. Conclusion

The diffusion welding of copper to aluminum was carried out and important factors for the joint strength were discussed. Results obtained are summarized as follows:

1) An intermetallic compound layer was observed which could be divided into three characteristic region I, II and III. The region I and III consisted of the intermetallic compound of θ ($CuAl_2$) and γ_2 (Cu_9Al_4), respectively. The hardness of region II was the highest of the regions.

2) The growing process of the intermetallic compound layer is considered to be controlled by the atomic diffusion and the increase in the intimate contact area between the faying surfaces. The atomic diffusion process became more important with the increase in welding temperature.

3) In the earlier stage of welding (shorter welding time), the tensile strength of joint increased with the rise of welding temperature, time and pressure. This stage is considered as a process where the intimate contact between the faying surfaces was developed.

4) In the later stage of welding (longer welding time), the tensile strength of joint welded at any temperature approached to a constant value (2 ~ 3 kg/mm^2) much lower than that of aluminum base metal. This tendency was observed for the thickness of the intermetallic compound layer greater than 15 ~ 20 μm. In this case, fracture was occured at the intermetallic compound layer, but not at the bonding interface. The strength of joint welded in the later stage is considered to be controlled by the strength of the intermetallic compound layer.

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