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Friction Welding of Silicon Carbide to Nickel with Intermediate Layer

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Abstract

An attempt has been made to apply the friction welding to the ceramics-to-metal joining with the aid of intermediate layers of active metals. The ceramics and metal specimens employed are a pressureless-sintered silicon carbide and a commercially pure nickel. The joint of SiC to Ni bonded without an intermediate layer were fractured without external load immediately after the bonding. Similarly, the joints bonded with the intermediate layers of zirconium and niobium were also fractured immediately after the bonding. On the other hand, the joint strength was improved pronouncedly by the use of the intermediate layer of aluminum. The maximum tensile strength of the joint with the aluminum intermediate layer reached about 130 MPa comparable with reported values of the strength of the diffusion-bonded and brazed joints. The dependence of the joint strength on welding parameters, friction time and forge pressure, is discussed on the basis of the observation of the microstructure and fracture morphology of the joint.

KEY WORDS : (Friction Welding) (Silicon Carbide) (Nickel) (Ceramics-Metal Joining) (Intermediate Layer) (Aluminum) (Zirconium) (Niobium)

1. Introduction

The friction welding is widely applied as a joining method of dissimilar metals¹⁾, since even a combination of dissimilar metals having very different properties (ex. hardness, melting point et al.) can be bonded without serious difficulty. However, to our knowledge, there have been only a few investigations reported of the friction welding of ceramics to metal²⁾. On the other hand, a number of investigations³⁾ have been reported of the application of the diffusion bonding and brazing to the ceramics-to-metal joining. One of the most generally accepted results obtained by them is that the joint efficiency of ceramics to metal can be improved pronouncedly by the use of the intermediate layer including active metal elements⁴⁾. In contrast to this, the application of the intermediate layer to the friction welding has been reported only in a very limited case⁵⁾, since a variety of dissimilar metal combinations can be friction-welded successfully by suitable choice of bonding parameters without an intermediate layer. However, considerable difficulty will be encountered in joining the combination of ceramics and metal even by the friction welding, since they have completely different properties. In the present investigation, therefore, the bond strength

and microstructure of the friction-welded joint of ceramics to metal have been studied with particular reference to the effects of the intermediate layer of active metals.

2. Experimental Details

A rod of SiC (silicon carbide) having excellent resistance to thermal shock was employed as the ceramics specimen, in view of the rapid heating and cooling rates of the heat-affected zone of the friction welding. The SiC specimen was produced by pressureless-sintering process with the densification aid of C. A commercially pure Ni (nickel), whose chemical composition is shown in **Table 1**, was employed as the metal specimen to be bonded to the SiC specimen, considering that there are many kinds of heat resistance alloys of nickel base. The dimensions of these specimens are shown in **Fig. 1**. The faying surface of the SiC specimen was finished by mechanical grinding, and that of the Ni specimen was finished to JIS 3S by turning in a lathe. Both faying surfaces were degreased in acetone immediately before the friction welding.

Active metal foils of Nb (niobium), Zr (zirconium) and Al (aluminum) were employed as the intermediate layer. The thickness of the Nb foil was 25 μm , that of the Zr foil 20 μm and that of the Al foil 30-1000 μm . These

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intermediate layers were placed between the ceramics and metal specimens as shown in Fig. 1. The chemical composition of the Al foil, with which joints standing the mechanical test and microstructure observation were obtained, is shown in Table 2.

The friction welding was carried out using the brake-type machine by pressing the Ni specimen (not rotated) against the rotated SiC specimen. The friction welding consisted of three stages as shown in Fig. 2: preliminary friction stage at preliminary pressure P_0 MPa for preliminary friction time t_0 s, friction stage at friction pressure P_1 MPa for friction time t_1 s and forge stage at forge pressure P_2 for forge time t_2 s. The preliminary friction stage was carried out in order to reduce the heavy impact load imposed on the friction-welding machine at the beginning of the contact. The forge pressure P_2 was applied just at the same time as the rotation was stopped. The values of these parameters employed are listed in Table 3.

The joint efficiency obtained was estimated from the tensile strength of the joint. The tensile test was carried out using an Instron-type machine at room temperature at a rate of $1.7 \cdot 10^{-3}$ mm/s. The device for the tensile test are shown in Fig. 3; i.e., the ceramics specimen was fixed to the device shown in Fig. 3 both by mechanical fastening with the screws and by adhesive. The joint was subjected to the tensile test without any machining after the welding.

Table 1 Chemical composition of the nickel specimen (mass%).

Fe	Cu	Pb	Mn	C	S	Co	Si	Ni
0.020	0.002	0.001	0.172	0.089	<0.001	0.012	0.045	Bal.

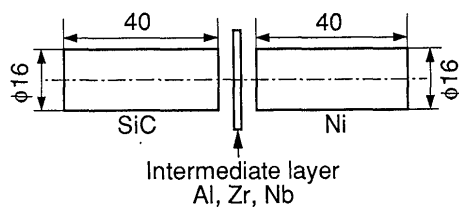


Fig. 1 Dimensions of the specimens used.

Table 2 Chemical composition of the aluminum foil for the intermediate layer (mass%).

Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Al
0.05	0.11	<0.01	<0.01	<0.01	<0.01	<0.01	<0.01	Bal.

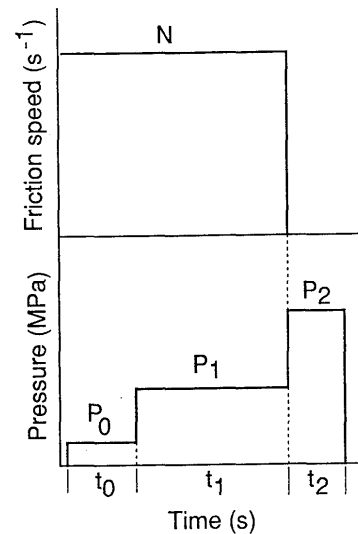


Fig. 2 Friction speed and pressure to the interface as a function of time during the friction welding.

Table 3 Parameters for the friction welding.

$N = 40 \text{ s}^{-1}$
$P_0 = 10 \text{ MPa}$
$P_1 = 20 \text{ MPa}$
$P_2 = 20 - 70 \text{ MPa}$
$t_0 = 1 \text{ s}$
$t_1 = 1 - 20 \text{ s}$
$t_2 = 6 \text{ s}$

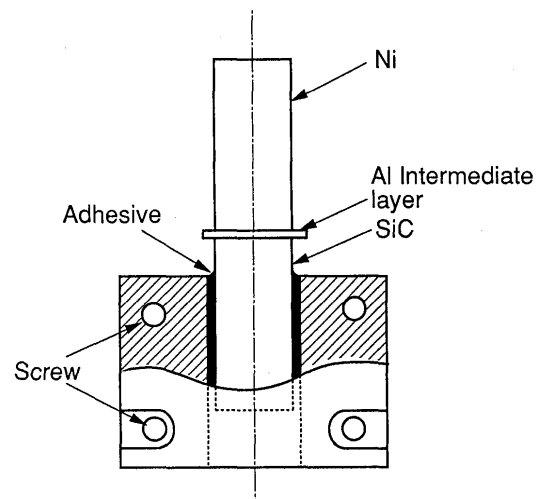


Fig. 3 Schematic illustration of the device for the tensile test of the joint.

3. Results and Discussion

When the SiC specimen was friction-welded to the Ni specimen without an intermediate layer, the joint was fractured without application of any external load

immediately after the bonding. Similarly, joints with the intermediate layers of Zr and Nb were also fractured immediately after the bonding. The fractured surfaces of these joints are shown in Fig. 4. All the fractured surfaces of the SiC side showed metallic colour rather than that of SiC itself, suggesting that a very thin metal layer was attached to the faying surface of the SiC specimen. On the fractured surface of the Ni side, small fragments of SiC were observed to be attached, though almost all of them were exfoliated during the cooling process after the bonding. These results suggest that a kind of bond, though very weak, was formed across the friction interface between the SiC and the metal surfaces.

On the other hand, when the aluminum intermediate layer was applied, joints having strength enough to stand the mechanical testing and microscopic inspection were obtained. Microstructures of SiC-to-Ni joints with the Al intermediate layer are shown in Fig. 5 for friction times from 2 to 20 s. A reaction layer was observed between the Al intermediate layer and the Ni specimen, with its thickness increasing with the friction time. On the other hand, no reaction layer could be observed between the SiC specimen and the Al intermediate layer, though some SiC particles were found to be embedded in the

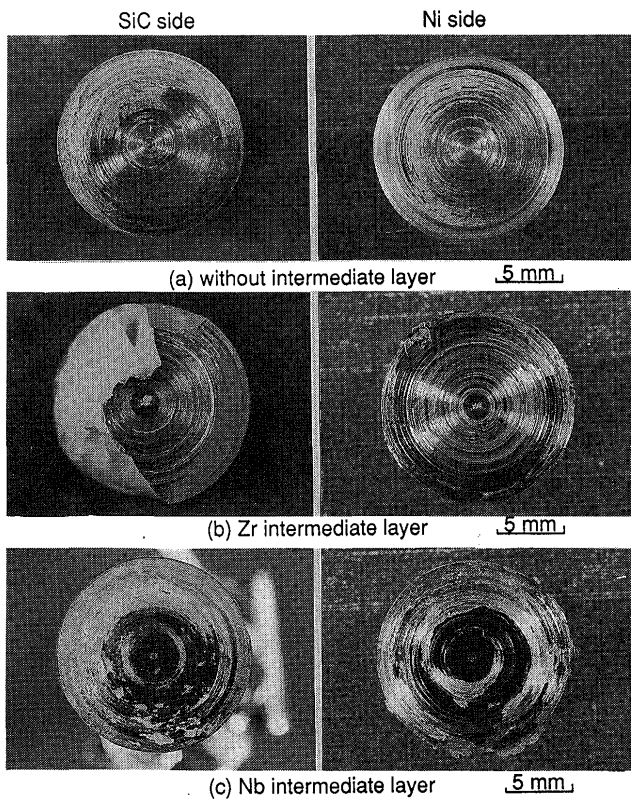


Fig. 4 Fractured surfaces of the friction-welded joints of SiC to Ni: (a) without an intermediate layer, (b) with the intermediate layer of zirconium and (c) with the intermediate layer of niobium.

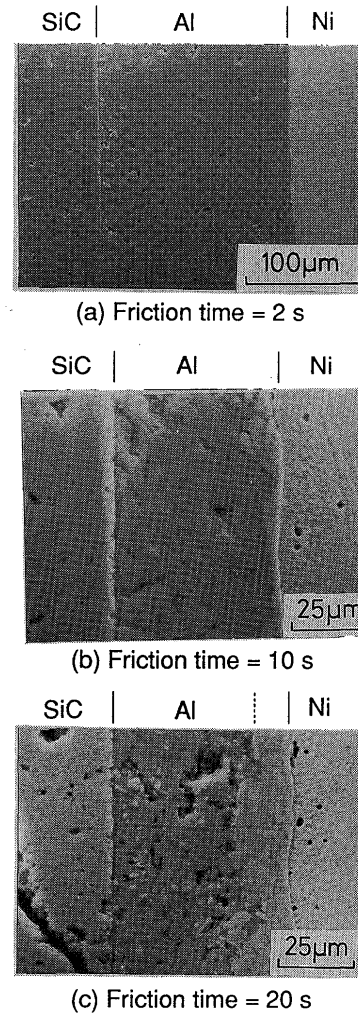


Fig. 5 Scanning electron micrographs of SiC-to-Ni joints with the intermediate layer of aluminum ($P_1 = P_2 = 20$ MPa).

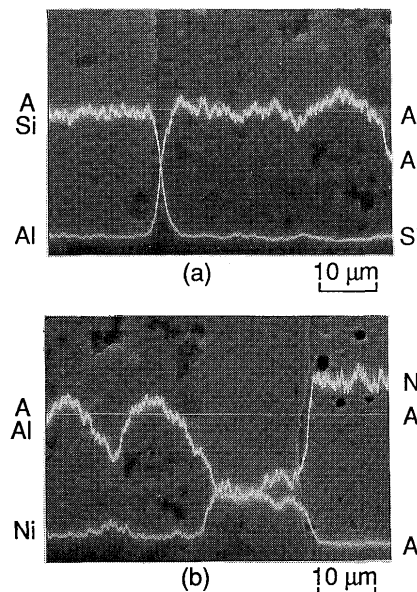


Fig. 6 Scanning electron micrographs of SiC-to-Ni joint with the intermediate layer of aluminum: (a) line analyses of Al and Si with EDX near the SiC-Al interface and (b) line analyses of Al and Ni near the Al-Ni interface ($P_1 = P_2 = 20$ MPa, $t_1 = 20$ s).

intermediate layer. These results were also supported by the concentration profiles of Ni, Al and Si as shown in Fig. 6. In the concentration profiles of Al and Si, no indication could be observed which suggested the formation of a reaction layer between the SiC specimen and the Al intermediate layer even for a friction time as long as 20 s, while terraces were observed in the concentration profiles of Al and Ni in the reaction layer between the Al and Ni. In order to identify the phases formed in the reaction layer, X-ray diffraction patterns from the fractured surfaces of joints were investigated using the Cu-K α line, since over a considerable area of the fractured surface cracks on tensile test propagated through regions close to the Al-Ni interface. As a result, diffraction lines identified as Al₃Ni and AlNi were detected on the fractured surfaces of both the SiC side and the Ni side. Therefore, the reaction layer between the Al intermediate layer and the Ni specimen can be considered to consist mainly of these intermetallic compounds. The thickness of the intermediate layer remaining at the bond was also influenced strongly by the friction time as shown in Fig. 7. The thickness of the remaining intermediate layer was decreased rapidly with increasing the friction time.

The effect of the friction time on the tensile strength of the joint is shown in Fig. 8. Though scattering widely, the joint strength exhibited a tendency to decrease with increasing the friction time. In order to explain this result, fractured surfaces of the joints were observed. As shown in Fig. 9(a), the fractured surface can be divided into region I, II, III, III' and IV according to their morphology and crack-propagating path. Results of closer observation of these regions are shown in Figs. 10, 11 and 12. Since both region I and region II presented dimple patterns (Figs. 10 and 11), ductile fracture in the Al

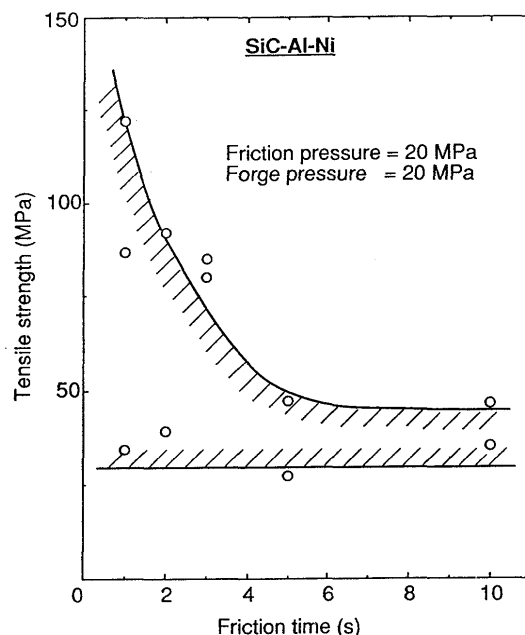


Fig. 8 Tensile strength of joints vs friction time for the friction welding of SiC to Ni with the intermediate layer of aluminum ($P_1 = P_2 = 20$ MPa).

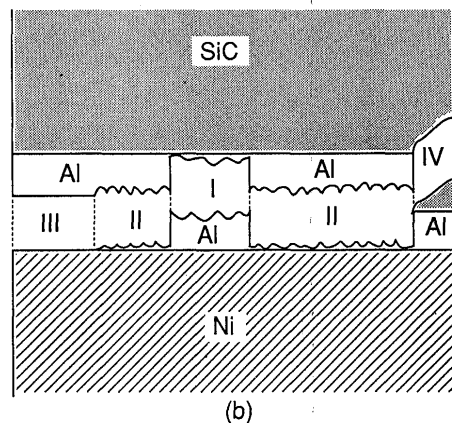
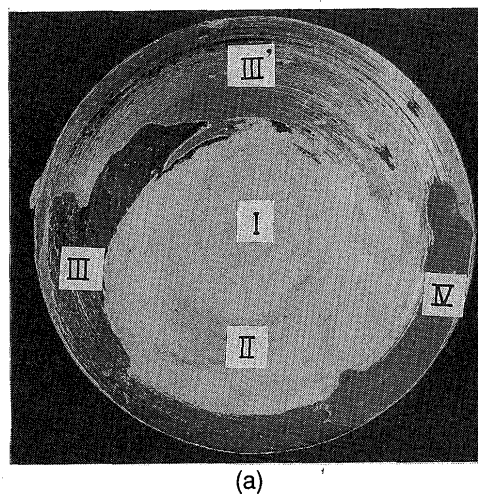


Fig. 9 (a) fractured surface of Ni side of a SiC-to-Ni joint with the intermediate layer of aluminum ($P_1 = P_2 = 20$ MPa, $t_1 = 10$ s), and (b) schematic illustration of the path of crack propagation.

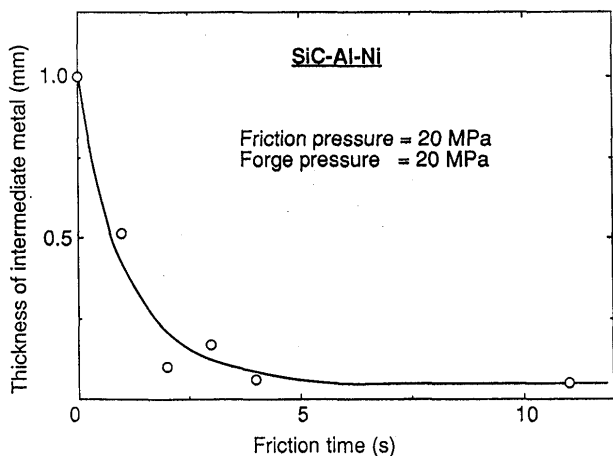


Fig. 7 Thickness of the aluminum intermediate layer remaining at the bond interface as a function of friction time ($P_1 = P_2 = 20$ MPa).

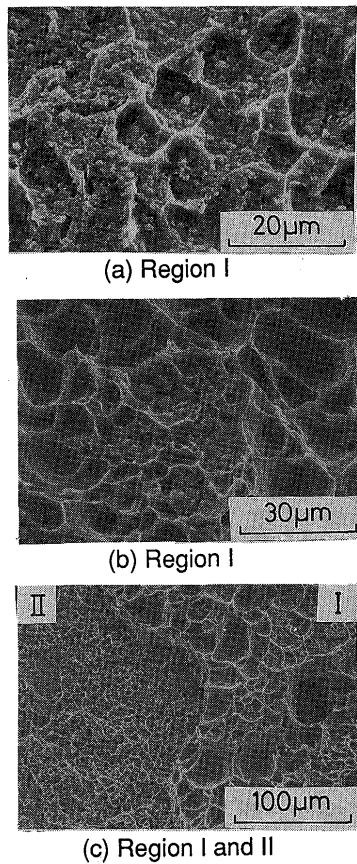


Fig. 10 Fracture morphology of regions I and II (Ni side): (a), (b) region I, and (c) boundary between regions I and II.

intermediate layer occurred in these regions. However, fracture location was different between these regions; i.e., fracture occurred in the Al intermediate layer close to the SiC-Al interface in region I and to the Al-Ni interface in region II as can be seen in Fig. 10. In addition, small particles of SiC were observed in region I, while no SiC particles could be observed in region II. Region III, presenting rather smooth and featureless morphology as shown in Fig. 11, can be regarded as the area where brittle fracture occurred at the Al-Ni interface with very slight load. Region III' is a part of region III where small islands of aluminum were attached to the fractured surface of the Ni side. In region IV, brittle fracture occurred in the SiC specimen as shown in Fig. 12. The path of the crack propagation suggested from these results is schematically illustrated in Fig. 9(b).

The fracture morphology of the joint was changed with friction time as shown in Fig. 13. When the friction time was short (the joint strength was higher), regions I and II showing ductile fracture morphology occupied almost all the fractured surface of the joint. In contrast, regions III and III' showing brittle fracture morphology were increased as the friction time was increased (the joint

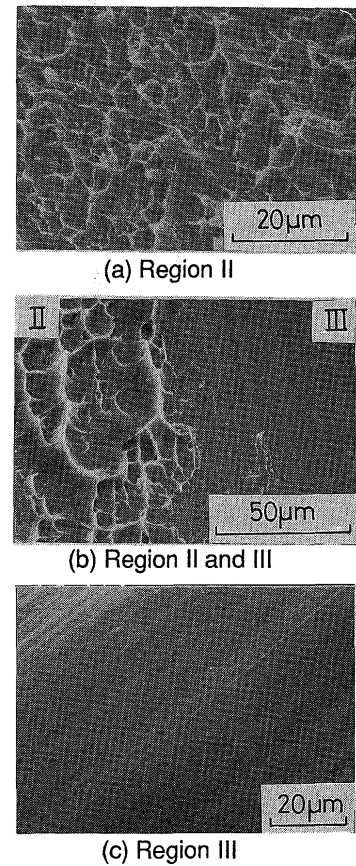


Fig. 11 Fracture morphology of regions II and III (Ni side): (a) region II, (b) boundary between regions II and III, and (c) region III.

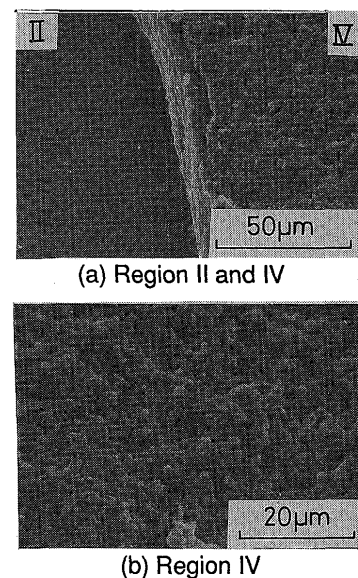


Fig. 12 Fracture morphology of regions II and IV: (a) boundary between regions II and IV, and (b) region IV.

strength was decreased). Therefore, the decrease in the joint strength with the increase in friction time can be related to the increase in the area of brittle-fractured regions III and III'.

On the other hand, the effect of the forge pressure on the tensile strength of joint was investigated in order to improve the joint efficiency. As shown in Fig. 14, the tensile strength of joint was increased with the forge pressure and reached about 130 MPa at a forge pressure of 70 MPa. The joint strength of 130 MPa is comparable with reported values⁶⁾ of the strength of joints obtained with the diffusion bonding and brazing. The fracture morphology of the joint was changed with forge pressure as shown in Fig. 15. As the forge pressure was increased, the area of region III was decreased, and that of region IV was increased; i.e., the area where fracture occurred in the SiC matrix was increased with forge pressure. This result suggests that the increase in the joint strength with forge pressure was resulted from the decrease in the area of region III.

Thus the tensile strength of the friction-welded joint of SiC to Ni with the Al intermediate layer was increased, as the friction time was decreased or as the forge pressure was increased. According to the observation of the fracture morphology of the joint, it seems that this

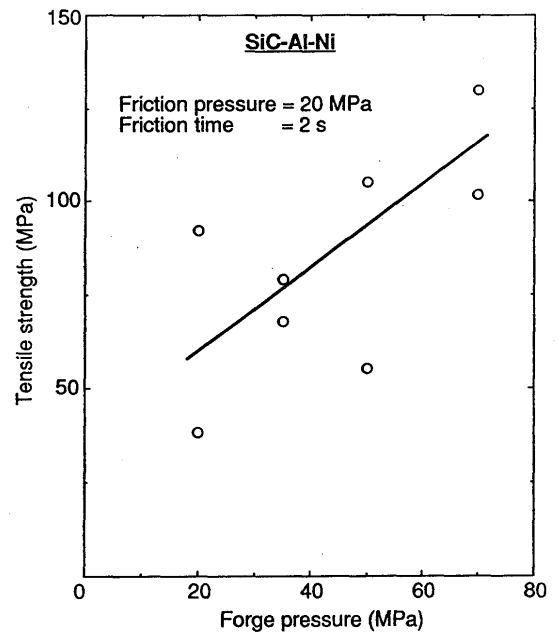


Fig. 14 Tensile strength of joint vs forge pressure for the friction welding of SiC to Ni with the intermediate layer of aluminum ($P_1 = 20$ MPa, $t_1 = 2$ s).

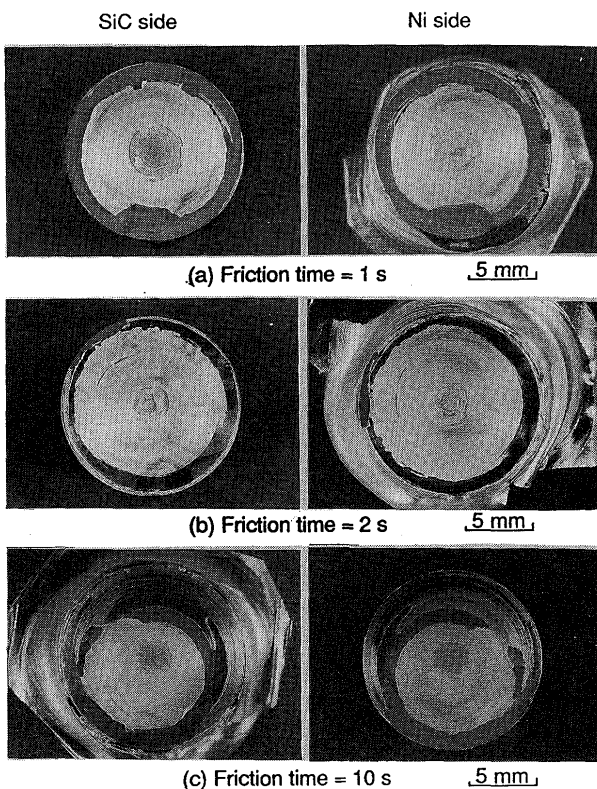


Fig. 13 Change in fracture morphology with friction time for the SiC-to-Ni joint with the intermediate layer of aluminum ($P_1 = P_2 = 20$ MPa).

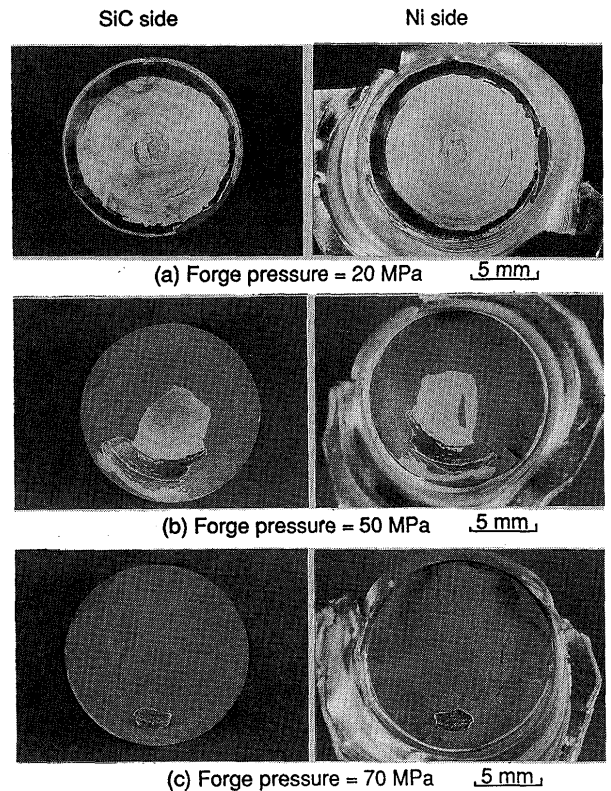


Fig. 15 Change in fracture morphology with forge pressure for the friction welding of SiC to Ni with the intermediate layer of aluminum ($P_1 = 20$ MPa, $t_1 = 2$ s).

increase in the joint strength was resulted from the decrease in region III where brittle fracture occurred at the Al-Ni interface. Important factors controlling the strength of the ceramics-to-metal joint can be listed as follows :

1. Residual stress generated during the cooling process after the bonding operation owing to the difference in the thermal expansion coefficient between ceramics and metal.
2. Formation of brittle interfacial reaction layer.
3. Presence of unbonded area or flaw.

Factor 1 is generally accepted as one of the most important factors in the diffusion bonding and brazing of ceramics to metal, and its effect on the joint strength becomes more serious as the thickness of the soft intermediate layer is decreased⁷⁾. This residual stress, however, usually causes the fracture in the ceramics matrix rather than the bond interface. In addition, as shown in Fig. 15, the area of region IV where fracture occurred in the SiC specimen was increased with the rise in the joint strength. These results suggest that the residual stress due to the difference in the thermal expansion is not an important factor controlling the joint strength of the friction welding of SiC to Ni with the Al intermediate layer.

Factor 2, the formation of brittle reaction layers, is also taken as a very serious problem in the joining of dissimilar materials. It has been said that the influence of the brittle reaction layer on the mechanical properties of the joint becomes weaker, as the thickness of the reaction layer is decreased, and is almost negligible at thickness less than few μm ⁸⁾. In the present investigation, the thickness of the reaction layer formed at the Ni-Al interface did not exceed $1\ \mu\text{m}$ for friction times less than 10 s. As shown in Fig. 13, however, regions III and III' still occupy considerable area of the fractured surface for friction times less than 10 s. Therefore, the formation of brittle reaction layers at the Al-Ni interface cannot be regarded as a main cause for the brittle fracture in regions III and III'.

Consequently, factor 3, the presence of unbonded area or flaw, seems to be responsible for the decrease in the joint strength with the increase in friction time; i.e., the unbonded area seems to be increased with the friction time. The reason for this can be related to the decrease in the thickness of the remaining intermediate layer (see Fig. 7) as follows. According to Hasui and Fukushima⁹⁾, the friction interface during the friction stage consists of two regions:

- (1) the central region of the interface where the intimate contact between the faying surface is attained over the whole region, and the material undergoes shear deformation without breaking, and

- (2) peripheral region where bonded and unbonded parts coexist with radial distribution, moving in the circumferential direction.

They have also shown that the area of the peripheral region consisting of unbonded and bonded parts was increased as the friction speed, namely the shear deformation rate, is increased. In our experiment, since the Al intermediate layer was much softer than the SiC and Ni specimens, the shear deformation was virtually carried out only in the intermediate layer. This means that the shear deformation rate at a given friction speed was increased as the intermediate layer became thinner. Since the remaining intermediate layer became thinner with increasing the friction time, it can be considered that the area of the peripheral region including unbonded parts at the friction interface was increased with friction time.

On the other hand, it has also been pointed out by Hasui and Fukushima⁹⁾ that the bonding over the whole friction interface can be achieved during the forge process through spreading-out of the central bonded region. In order to spread out the bonded region, of course, materials must be supplied to unbonded parts by the plastic flow in the radial direction of the friction interface. This plastic flow in the radial direction is interfered with by the plastic constraint effect from the harder Ni and SiC specimens, as the thickness of the Al intermediate layer is decreased. Thus, it is likely that as the friction time was increased (as the thickness of the remaining intermediate layer was decreased), it was more difficult to spread out the central bonded region and eliminate the unbonded part.

Therefore, it can be concluded that the strength of the friction-welded joint of SiC to Ni with the Al intermediate layer was decreased with increasing the friction time for two reasons resulting from the decrease in the thickness of the remaining intermediate layer:

- (i) the increase in the area of the peripheral region including considerable unbonded parts during the friction stage, and
- (ii) the plastic constraint effect from the Ni and SiC specimens which interfered with spreading out the central bonded region during the forge stage.

The effect of the increase in the forge pressure was probably to spread out the central bonded region to a more extent. This effect can be regarded as the cause for the increase in the joint strength with the forge pressure (see Fig. 14).

4. Conclusions

An investigation has been made of the effects of active-metal intermediate layers on the friction bonding of

a pressureless-sintered silicon carbide to a commercially pure nickel. The intermediate layers used were the foils of zirconium 20 μm thick, niobium 25 μm thick and aluminum 30-1000 μm thick. When the intermediate layers of Zr and Nb were applied, joints were fractured without any external load immediately after the bonding similarly to those bonded without an intermediate layer. On the other hand, the joint strength was improved pronouncedly by the use of the intermediate layer of aluminum. The maximum tensile strength of the joint with the Al intermediate layer was about 130 MPa comparable with that of the diffusion-bonded and the brazed joints. The tensile strength of joints with the Al intermediate layer was increased as the forge pressure was increased or as the friction time was decreased. From the observation of the fracture morphology and microstructure of the joint, it was suggested that the decrease in the joint strength with increasing the friction time could be attributed to the increase in the

uncontacted area at the friction interface. The effect of the increase in the forge pressure was probably in increasing the bonded region at the friction interface.

References

- 1) J. Hassler et al.: *Welding Handbook 3A 6th ed.*, MACMILLAN, London, (1970), p.50.1.
- 2) A. Suzumura, T. Onzawa, Y. Arata, A. Ohmori and S. Sano: *J. High Temp. Soc. Jpn.*, **13-1**(1987), 43 (in Japanese).
- 3) M. G. Nicholas: *Advanced Joining Technologies* ed. by T. H. North, Chapman and Hall, London, (1990), p.160.
- 4) H. Mizuhara and K. Mally: *Weld. J.*, **64-10**(1985), 27.
- 5) S. Elliott and E. R. Wallach: *Metal Const.*, **13-4**(1981), 221.
- 6) T. Yano, N. Takada and T. Iseki: *J. Ceram. Soc. Jpn.*, **95-3**(1987), 375 (in Japanese).
- 7) K. Hamada, M. Kureishi, M. Ueda, T. Enjo and K. Ikeuchi: *Q. J. Jpn. Weld. Soc.*, **3-3**(1985), 483 (in Japanese).
- 8) S. Elliott and E. R. Wallach: *Metal Const.*, **13-3**(1981), 167.
- 9) A. Hasui and S. Fukushima: *J. Jpn. Weld. Soc.*, **44-12**(1975), 1005 (in Japanese).