

REACTIVE SINTER BRAZING OF PM ALUMINIUM TO SINTERED AND WROUGHT STEELS

1. FILLER SELECTION AND PROCESSING

H. Danninger, Y. Wang, E. Wolfsgruber, J. Seyrkammer

Abstract

For joining of PM components consisting of Al alloy Al-Cu-Mg-Si to wrought steel, brazing during sintering of de-waxed Al compacts was studied. Direct reaction between Al and Fe does not occur, and a liquid phase is necessary to trigger off the reaction. While commercial braze fillers used for joining wrought Fe and Al were found to be less suitable, Cu base fillers pressed from powders resulted in sound joints by forming liquid phase with Al through contact melting and subsequent reaction to form aluminides, although external force by application of loads is necessary to obtain the required contact. Cu fillers containing elements that react with Al exothermically are particularly attractive, the reaction apparently further boosting the joining process. Intermetallic Al-Fe phases are formed inevitably at the joint but do not cause major problems, fracture during mechanical testing occurring frequently within the Al matrix rather than in the joint. Sintering temperature and amount of filler added however have to be carefully optimized in order to obtain satisfactory joining while avoiding excessive shrinkage of that part of the Al matrix that is in close contact with the melt, which would mechanically weaken this area and lead to intolerable distortion.

Keywords: *sintered components, PM aluminium, sinter brazing, reactive brazing*

INTRODUCTION

Weight reduction is a primary target in automobile production, the reasons being the demand for lower fuel consumption and less emission of exhaust gases. Every single component has to be checked for possible reduction of weight, and this holds also for the precision parts produced by powder metallurgy techniques. Weight reduction may be attained by using aluminium in place of steel, and PM aluminium components have increasingly found their market for automotive applications [1-3], although the tonnage of PM Al parts is still small compared to that of steel components.

Frequently, the optimum solution for a PM part is manufacturing involving a joining process. This may be due to geometry reasons, i.e. two or more simple-geometry parts are separately manufactured and then joined to form a complex shape that is not accessible through pressing in one step. Another reason may be that the combination of different materials offers specific advantages, e.g. by offering specific properties only on

those places where it is actually needed. This may also hold for PM Al components which have the advantage of low weight and inertia but on the other hand have low strength, ductility, and hardness which properties in turn are offered by PM or wrought ferrous materials. Joining of PM Al alloys and sintered or wrought steels therefore seems to be an interesting topic.

For joining of PM parts, numerous techniques are possible (e.g. [4-6]), mostly corresponding to those well known from classical metallurgy, although some specialties of powder metallurgy are well established today such as e.g. sinter bonding. For sintered Al, in particular its low ductility must be considered. This might be a disadvantage e.g. for press-fitting if there are tensile loads on the Al; similarly, sinter bonding, which is well suited for concentric parts, might be difficult if the outer part is Al since Al tends to exhibit higher expansion than the inner, ferrous one. Special welding techniques such as laser welding might be applicable but require expensive equipment. The most promising techniques seem to be sinter brazing and adhesive bonding, sinter bonding being also possible in the case of appropriate design. The principal difficulties expected were obtaining sufficiently strong joints, especially in the case of metallic bonds being developed, since formation of brittle intermetallic phases is inevitable in this case. Furthermore, maintaining the required dimensional tolerances was regarded not simple, too, and most probably the joined assembly would have to be sized (but Al precision parts have to be sized anyhow).

The work concentrated on sinter brazing, which seems to be economically most attractive, after sinter bonding was regarded not feasible as will be shown here. For optimizing the filler material, model specimens were prepared and tested. Subsequently, the developed process was employed on a combination of Al camshaft belt pulleys fitted with wrought steel hubs as will be described in Part II of this work.

EXPERIMENTAL PROCEDURES

For optimizing the process and in particular the filler material used for brazing, model assemblies were produced from test bars made of Al and PM iron, respectively (Fig.1). In order to avoid formation of angled specimens during sinter brazing (which would result in unpredictable stresses during testing), the joining was done with intact tensile test bars, and after sinter brazing the Al bars were cut to the joint using a fine saw. Thus, exactly straight specimens were obtained that could be properly measured under tensile load.

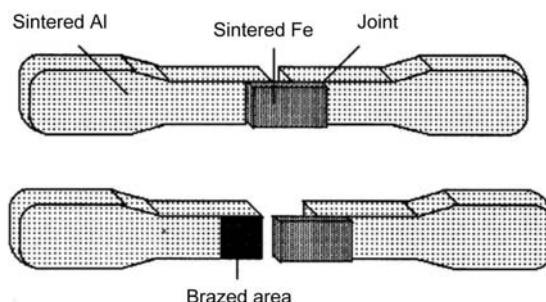


Fig.1. Model assembly for principal joining tests and measurement of shear strength.

PM iron bars were produced from atomized iron powder ASC 100.29. The powder was mixed with 0.5 mass% pressing lubricant and compacted to rectangular bars of 60 x 7 x approx 6 mm³. The green compacts were sintered 30 min at 1120°C in a pusher furnace with Mo heating elements, the atmosphere being hydrogen of technical purity.

The Al bars were produced according to the guidelines described in [7,8]. Alumix 123 (Ecka Granulate GmbH, approx. Al-4.5% Cu-0.5% Mg-0.5% Si) containing 1.5 mass% lubricant was compacted at 300 MPa to standard tensile test bars (ISO 2790). The specimens were de-waxed in a laboratory pusher furnace at 400°C in flowing nitrogen of 99.999% purity.

Fillers were either prepared by cutting rectangular pieces out of sheet material or by pressing thin bars of typically 60 x 7 x 0.2 mm from powder mixes, no lubricant being used here.

Sintering (in fact sinter brazing) of the model specimens was done in a small pusher furnace operated under high purity nitrogen. De-waxing and sintering/sinter brazing was preferably done in separate runs. If de-waxing and sintering was done in one single run, the wax residues tended to react with the iron, resulting in oxidized and discoloured parts, as described below. Furthermore, earlier work in PM Al camshaft belt pulleys [8,9] had already shown that separate de-waxing is recommendable, especially for heavy, thick-walled components.

SINTER BONDING

First of all it was checked if the Al compact can be pressed directly on top of a sintered ferrous part. This joining technique has been successfully employed e.g. for composite rocker arms [10]. For joining Al and Fe, an additional positive effect might be the exothermic reaction between Al and Fe that should boost joining. Thus, flat sintered Fe bars 60 x 7 mm with approx. 3 mm thickness were inserted into the 60 x 7 mm pressing tool, Alumix powder was filled on top and pressed at 300 MPa. The composite compacts were then sintered in the laboratory pusher furnace (de-waxing 30 min, 400°C, sintering 30 min, 600°C, high purity nitrogen). The joints were investigated metallographically.

It showed (Fig.2) that metallic bridges, consisting of intermetallic phase, had been formed only locally. Between those bridges, crack like pores remained, indicating inhibition of the reactive sintering probably by the oxide layer on Al. It can be assumed that the local bridges are formed only if Cu particles are present between Al and Fe, resulting in transient liquid phase that can penetrate the oxide layers [11,12].

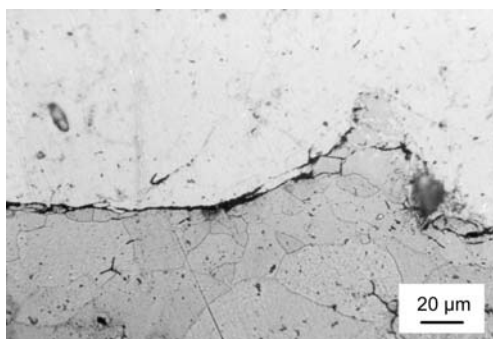


Fig.2. Joint of Alumix 123 powder pressed onto PM iron and then sintered.

There might however be a geometrical effect, too, inhibiting joining. It is well known that also for ferrous parts, pressing of powder on top of a flat sintered part does not result in satisfactory interface strength. Profiled surfaces are superior, as shown in [10]. Therefore, also here Fe bars with saw tooth profiled surfaces were tested. It showed that in fact the joint is much better than in the case of flat Fe surface. However, the problem with staining of Fe, apparently due to wax residues, remained. It was therefore decided to proceed with joining of Fe and de-waxed Alumix compacts, which of course excludes pressing Al powder mix on sintered iron.

In this case, brazing seems to be attractive since the tests described above had indicated that liquid phase is helpful to trigger off the bonding reaction to result in satisfactory joints (in that respect, sintering of Alumix can be termed also as "micro brazing" of Al particles). Sinter brazing, i.e. including the joining process in the sintering of the Al pulleys, was economically most interesting. For this technique, suitable filler materials had to be found.

BRAZING WITH LOW-MELTING FILLERS

For joining of wrought Al and steel, filler materials are commercially available. First brazing tests were thus carried out with a filler paste specially recommended for Al/Fe joining (E+C Super fluid 190 Al, supplied by Castolin). Basic sintering tests were carried out using models as depicted in Fig.1; as recommended by the filler manufacturer, joining was done at 590°C for 30 min. Both de-waxed and as-sintered Alumix was used. The tests showed however that this filler does not result in satisfactory joints; very weak bonding was observed, and in parts no bonding at all. This was attributed to the organic binder phase of the paste that probably enters the pores during heating and during sinter brazing in part decomposes and contaminates the interfaces. This filler seems to be suited for wrought materials but not for porous structures. Thus it was concluded that special filler materials had to be developed.

Filler materials for successful brazing should either melt below the sintering/brazing temperature - i.e. below approx. 600°C - or form liquid phase through reaction with one of the components, i.e. by contact melting. Among the first group, tin and zinc can be mentioned; Zn is a standard element in 7xxx type Al alloys (e.g. [13]), forming transient liquid phase [14,15] while Sn has been shown to result in persistent liquid phase during sintering of Al alloys [16]. In the second group especially Cu is noteworthy since it also enables sintering of Al materials by contact melting [12]. Of course also low-melting eutectic compositions - Al-Si, Al-Cu, Al-Mg - might be useful since Al-Si is largely used for brazing Al/Al.

Experiments were carried out with Sn base and Zn base fillers. Powders were mixed and pressed to thin sheets. Assemblies as shown in Fig.1 were sinter brazed in the laboratory pusher furnace. For Sn base fillers, also soldering at 300°C was tried; in this case of course fully sintered Al compacts were used.

It showed that soldering with Sn at 300°C is not suitable, due to insufficient wetting of Al. Brazing at temperatures between 580 and 620°C however resulted in metallic joints especially in the case of Sn base fillers, wetting being observed although no Mg was contained in the filler [17]. Apparently the Mg present in the Al compacts affords sufficient wetting. The best results were obtained with Sn-Cu fillers containing 20 to 50% Cu; here, also better strengthening can be expected compared to plain Sn fillers [17]. In the case of Cu content <20%, pronounced porosity is found in Al adjacent to the joint (Fig.3). At high Cu contents (>50%) no bonding at all was observed. Also in the case of satisfactory

brazing, however, pronounced shrinkage of the Al matrix was encountered near the joint, apparently due to the liquid phase present here (Sn is insoluble in Al and thus forms persistent liquid phase). This shrinkage might offer problems with dimensional stability when brazing precision components.

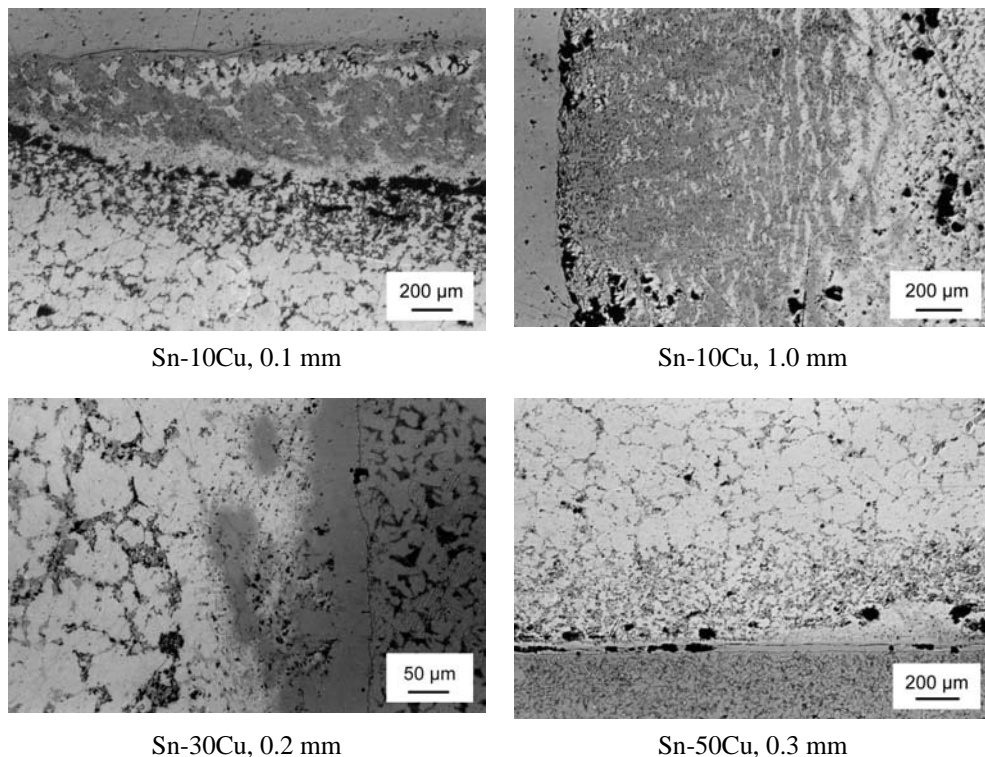


Fig.3. Section of PM Alumix/PM Fe joint, sinter brazed using as filler Sn-Cu compacts with varying compositions and layer thickness.

Zn base fillers proved to be less suitable than Sn based ones; the joints obtained were weak and contained many pores. Furthermore, pronounced swelling of Al near the joint was observed. This is due to the high solubility of Zn in Al, which is also responsible for the intermediate expansion and frequent dimensional problems encountered during sintering of Al-Zn(-Mg) compacts [18,19].

Low-melting eutectic fillers were also tested. It showed that Al-12% Si (typical filler for Al brazing) bonds well to Al but not to iron, which was surprising since it was expected that the filler would exothermally react with Fe to form iron aluminide. Al-Mg fillers (33 to 60% Mg) on the other hand did not bond to Al but to some extent to Fe. Al-33% Cu in contrast resulted in bonding to both partners and thus in fairly satisfactory joints.

BRAZING WITH CONTACT-MELTING FILLERS

It was to some extent surprising that Sn-Cu fillers containing >50% Cu had not resulted in good joining at temperatures near 600°C since the Al-Cu eutectic temperature is about 550°C and thus markedly lower than the actual sintering temperature. I.e. even for

plain Cu fillers, liquid Al-Cu eutectic should be formed through contact melting, and the melt then should react with iron by forming aluminides.

Experiments with standard Cu sheet material used as filler showed that also in this case, joining does not occur since there is apparently no liquid phase generated. This was attributed to insufficient mechanical contact between the Cu sheet and the Al compact, the oxide layer that covers Al in fact preventing formation of metallic contacts between Al and Cu that are essential for contact melting. (In pressed Al-Cu mixtures the contact between the two metals is of course much closer).

Kehl [20] had shown that for Al-X systems, contact melting can be enhanced by mechanical loading of the contact zone. Thus, further tests were carried out with external loads being put on top of the sandwich specimens. In fact it was found that an external loading of approx. 0.02 N.mm^{-2} results in the expected contact melting between Al and Cu and in formation of excellent joints that contained only a few defects. Also the effect on the Al matrix near the joint was much less than with Sn, although also here increased porosity can be found. A typical joint is shown in Fig.4; slight penetration of the iron by Cu to approx. $200 \mu\text{m}$ depth is discernible. There is a double layer of intermetallic phase at the interface (Fig.4b); formation of aluminides is inevitable with Al-Fe joints, and the effect of these brittle phases of course has to be investigated.

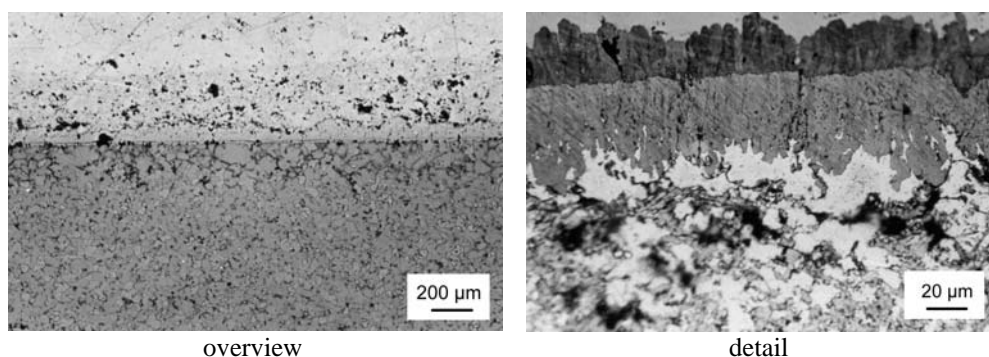


Fig.4. Section of PM Alumix/PM Fe joint, sinter brazed using Cu sheet as filler.

Cu thus seems to be a suitable filler material for sinter brazing of Al and Fe. It was however expected that further activation of the brazing process should be additionally beneficial. Brazing might be boosted by introduction of components into the filler that react with Al exothermally during contact melting, i.e. in particular Fe and Ni. Local overheating was thus expected to occur and possibly formation of defect-free joints. Of course, Fe reacts with Al in the course of the brazing process anyhow, but offering some Fe or Ni already at the onset of melt formation might accelerate the bonding process. Lower brazing temperatures might also be possible in the case of reactive fillers but were regarded unattractive because of adverse effects on the sintering of Al, low sintering temperatures resulting in expansion and insufficient strength [7,8]. Therefore, this alternative was regarded feasible only when joining already sintered Al parts.

Since Cu-Fe and Cu-Ni sheet material was not commercially available, filler materials were prepared from Cu base powder mixtures. Electrolytic Cu powder was mixed with the additive powder - in particular carbonyl Fe and Ni grades - and compacted at 400 MPa to form very thin bars of desired thickness. Sandwich specimens were prepared as described above, and sinter brazing was done in high purity N_2 . It showed that the reactive

fillers resulted in significantly better joints compared to plain Cu sheet, the joints being virtually defect-free, and also the adjacent Al matrix was less affected than in the case of Cu, hardly any increased porosity being discernible compared to e.g. Cu-Sn fillers (Fig.5a, b).

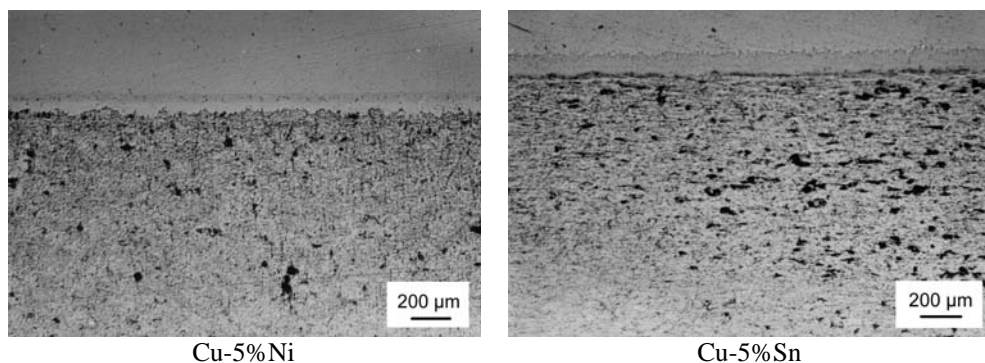


Fig.5. PM Alumix / PM iron sinter brazed at 600°C; Cu base fillers.

The joints were also studied by microprobe analysis. In Figures 6a, b, secondary electron images are shown; it is clearly visible that there are 4 intermediate layers between the iron and Al matrices, respectively. When studying the joint from plain Fe to Al, the fairly thin (approx. 50 μm) double layer of intermetallic phase immediately at the contact zone is followed by a coarse two-phase layer of 100 to 150 μm thickness in which equiaxed grains of dark (Al-rich) and light phase are rather evenly mixed. In the much thicker – about 1 mm – adjacent zone, light phase forms fine eutectic-like structures within the dark matrix.

EDS point analysis and X-ray mapping showed (Fig.6c-f) that the mentioned double layer of intermetallic phase consists of a binary phase at the Fe side which contains Al and Fe in an atomic ratio of approx. 75% to 25% (and thus is apparently Al_3Fe) and a ternary phase on the Al side that exhibits a composition of approx. 70 at% Al-20 at% Cu-10 at% Fe (with small amounts of Ni). According to the ternary phase diagram Al-Cu-Fe given in [21] this should be the ternary ω phase. The darker spots found in the second, ternary layer contain mostly Al, Ni, and Fe with only small amounts of Cu (in at% approx. 82:9:7:2).

The coarse two-phase layer contains considerable amounts of Ni, as shown by the x-ray image for Ni $K\alpha$. EDS analysis showed the layer to consist of an Al phase with small amounts of Cu (approx. 3 wt%) and a second phase - white in the SE image - with a composition (at%) of approx. 70Al-20Cu-10Ni. Within the finely structured broad zone the matrix is more or less identical but contains binary Al-Cu phase. The eutectic-like structure indicates that in this region simply liquid phase sintering of Al-Cu with high Cu content has occurred, apparently due to the larger amount of melt from the joint. This indicates that Cu is predominantly consumed by the Al matrix, the Al-Cu melt formed during sinter brazing being drawn into the pores of the Al compact by capillary forces.

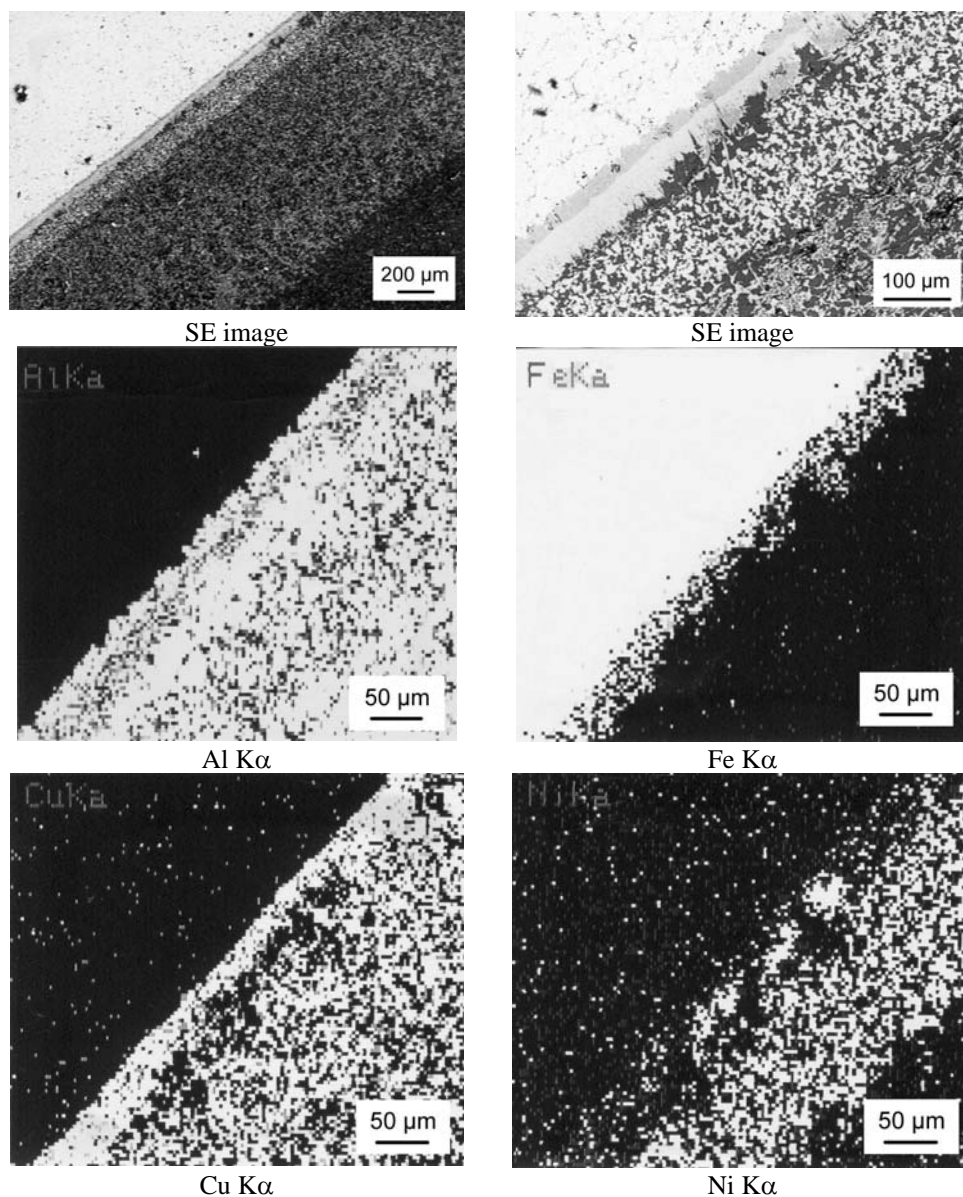


Fig.6. SEM micrographs of PM Alumix / PM iron joint, Cu-10Ni filler.

It can thus be concluded that the liquid phase generated by contact melting causes formation of intermetallic phases - aluminides - at the direct interface iron-Al. The liquid also penetrates the Al matrix to a given depth, causing pronounced liquid phase sintering. During cooling, Fe and Ni containing aluminides are precipitated near the interface while in the remaining infiltrated zone the Al-Cu melt solidifies and forms eutectic structures of Al-Cu phase in the Cu saturated Al matrix.

MECHANICAL TESTING

The structure of the joint at first seems to be far from optimal concerning the mechanical strength. The intermetallic phases must be expected to cause embrittlement of the joint and fracture in particular in the case of impact loading. Mechanical testing thus was regarded essential.

The mechanical strength of the joints was measured by shear testing, i.e. by tensile testing of joined specimens using specimens as shown in Fig.1. Frequently, fracture did not occur within the brazed layer but either in the adjacent Al matrix or simply across the Al bar, as schematically shown in Fig.7.

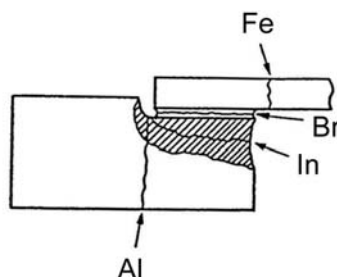


Fig.7. Fracture paths observed during testing of sinter brazed Al-Fe assemblies (schematically). Fracture paths: Al: base Al, In: Al infiltrated by filler, Br: joint; Fe: base iron.

The fillers tested included not only Cu, Cu-Fe, and Cu-Ni, but also other binary systems were studied. The shear strength values obtained are given in Table 1 and graphically depicted in Fig.8; in the Table also the respective fracture path is indicated, and those strength values that cannot be referred to fracture through the joint, i.e. to the defined brazed area, are given in *italics*.

Tab.1. Mechanical strength of Alumix 123/PM iron joints Al: Compacted 300 MPa, dewaxed; Fe: compacted 400 MPa, sintered 30 min 1120°C in H₂. Assembly sinter brazed 30 min at 600°C in N₂. Thickness of filler is 0.25 mm.

Filler composition [mass%]	Shear strength [MPa]	Path	Filler composition [mass%]	Shear strength [MPa]	Path	Filler composition [mass%]	Shear strength [MPa]	Path
Cu	12.0	Br	Cu-2Ni	25.5	In	Cu-2Sn	28.0	Al
Cu-2Fe	27.6	In	Cu-5Ni	37.6	In	Cu-5Sn	32.1	Al
Cu-5Fe	34.1	In	Cu-10Ni	33.5	In	Cu-10Sn	27.6	In
Cu-8Fe	46.0	In	Cu-0.5Cr	27.8	In	Cu-15Sn	27.7	In
Cu-10Fe	28.4	In	Cu-2Cr	37.8	In	Cu-20Sn	21.1	In
Cu-12Fe	21.4	In	Cu-5Cr	33.1	In	Cu-30Sn	12.4	Br
Cu-0.5Mo	22.0	Br	Cu-0.5Ti	16.6	Br	Cu-5Zn	3.5	Br
Cu-2Mo	25.7	In	Cu-2Ti	19.9	Br	Cu-1Mg	5.6	Br
Cu-5Mo	30.8	In	Cu-5Ti	13.0	Br	Cu-2Mg	7.6	Br

As can be seen from Figure 8, the highest (nominal) shear strength levels are attained by Cu-X fillers containing Fe and Ni, i.e. elements that exothermically form intermetallic phases. Also Cr and Mo proved to be suitable additives while Ti, which is

contained in some reactive braze fillers, is less recommendable, although it also generates a very exothermic reaction when forming aluminides. Here it may be supposed that the sensitivity of Ti not only to oxygen but also to nitrogen is a disadvantage, Ti reacting with interstitials before reaction with Al is possible. A similar effect may be responsible for the poor results with Cu-Mg fillers, also Mg being a very oxygen sensitive metal (prealloyed Cu-Mg material may be less critical here). Cu-Sn fillers do not result in quite as strong joints as do e.g. Fe and Ni but are attractive in practice since bronze sheet material is easier to obtain commercially than Cu-Fe compacts.

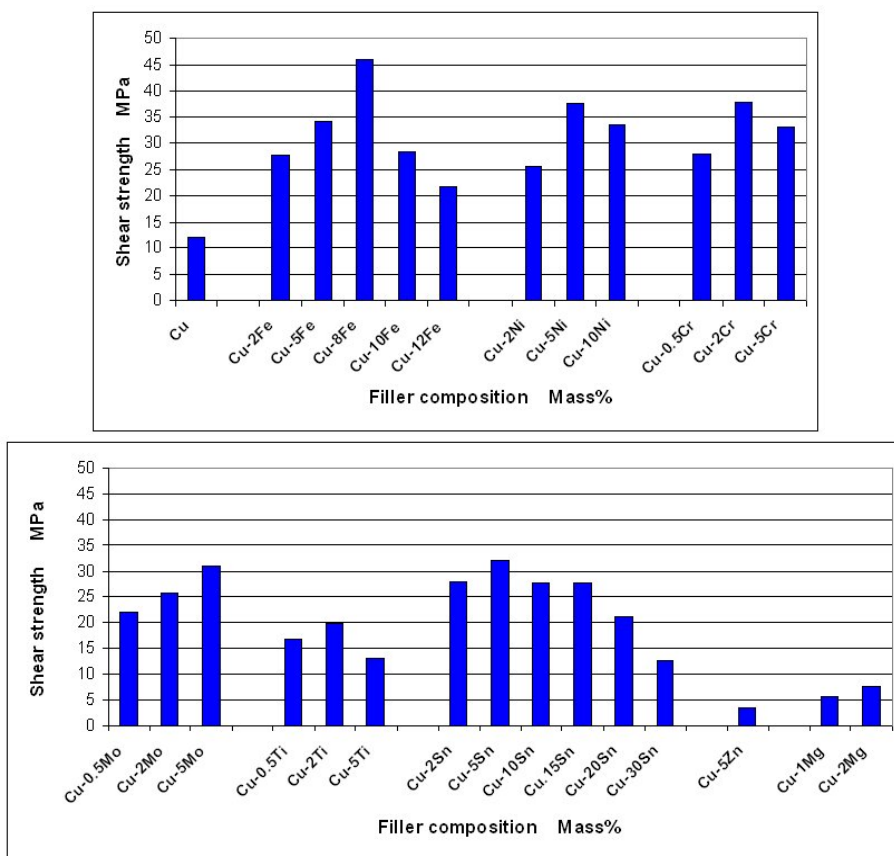


Fig.8. Mechanical strength of Alumix 123 / PM iron joints, sinter brazed at 600°C using different fillers.

Generally it can be stated that the shear strength of the joints is only moderate, being considerably lower than the strength of the Al matrix. This at first would surely be attributed to the embrittling effect of the intermetallic phases generated at the interface. However, as can be seen from Table 1, fracture in most cases did not occur along the interface, except in those cases when metallic bridges were formed only locally and pores were found within the joint. In particular if sound joints were obtained, fracture occurred rather within the part of the Al matrix adjacent to the joint, which had been infiltrated by the liquid phase.

Sections had shown that the liquid phase occasionally results in formation of coarse pores - probably through Ostwald ripening, see [22], and these pores affect the strength of the transition zone. Cu-Fe and Cu-Ni filler materials resulted in less pronounced pore coarsening than plain Cu, and not surprisingly the shear strength attained is also considerably higher. Thus the low strength of the joint is not so much caused by the aluminides, which are surely brittle but also hard and strong (microhardness HV0.05 in excess of 1100), but rather by the weakening of the adjacent Al matrix.

In principle the moderate shear strength of the Al-Fe joint might be expected to be a disadvantage. However it must be considered that in many cases, by appropriate geometry of the joint - e.g. with a concentric joint square or hexagonal in place of circular - most of the load can be transferred by mechanical interlocking, and the metallic joint acts only as a safety measure holding both partners in place. Furthermore, Al-Cu-Mg-Si sintered components are mostly used when geometrical precision and low weight, rather than superior mechanical strength, are required, and therefore the strength properties of the brazed joint should not pose too much of a problem.

CONCLUSIONS

For joining dewaxed Al-Cu-Mg-Si powder compacts to wrought and PM ferrous materials, reactive sinter brazing showed to be a successful way. Al and Fe do not react directly but a liquid phase is necessary to trigger off the reaction forming iron aluminides. Cu base filler materials generating Al-Cu melt proved to be viable, resulting in sound joints with adequate mechanical strength if external load was applied. At the interface, intermetallic phases are formed through reaction of Al and Fe promoted by the Cu base melt, usually a double layer of a binary iron aluminide and a ternary Al-Fe-Cu phase being found. In the adjacent Al matrix, Al-Cu phases are precipitated; this part of the matrix, which is infiltrated by the liquid filler, is definitely the weakest zone of the joint. Cu-Fe and Cu-Ni fillers are better than plain Cu, the Al matrix adjacent to the joint being less affected than in the case of plain Cu. Generally the shear strength of the joint is moderate, values of about 30-45 MPa being obtained with sound joints. Suitable design of the joint geometry, making use of mechanical interlocking, is therefore recommended for load bearing components.

REFERENCES

- [1] Neubing, HC. In: Pulvermetallurgie in Wissenschaft und Praxis, vol. 20. Compiled by H.Kolaska. Hagen : Fachverband Pulvermetallurgie, 2004, p. 3
- [2] Delarbre, P., Krehl, M. In: Proc. Conf. Aluminium & Light Alloys for Automotive Applications. Princeton : Troy MI, MPIF, 2000, p. 33
- [3] Kruzhanov, V., Neubing, HC. In: Pulvermetallurgie in Wissenschaft und Praxis, vol. 19. Compiled by H.Kolaska. Hagen : Fachverband Pulvermetallurgie, 2003, p. 177
- [4] Boyer, HE. In: Metals Handbook, vol. 7. 9th ed. Materials Park : ASM, 1984, p. 451
- [5] Koch, HP., Pohl, D. In: Pulvermetallurgie in Wissenschaft und Praxis, vol. 9. Compiled by H.Kolaska. Düsseldorf : VDI-Verlag, 1993, p. 36
- [6] Ratzi, R. In: Pulvermetallurgie in Wissenschaft und Praxis, vol. 18. Compiled by H.Kolaska. Hagen : Fachverband Pulvermetallurgie, 2002, p. 187
- [7] Neubing, HC., Jangg, G.: Metal Powder Rep., vol. 42, 1987, no. 5, p. 354
- [8] Jangg, G., Danninger, H., Schröder, K., Abhari, K., Neubing, HC., Seyrkammer, J.: Materialwiss. u. Werkstofftech., vol. 27, 1996, p. 179
- [9] Danninger, H., et al. In: Adv. Powder Metall. & Partic. Mat. 1992, vol. 6. Compiled by J.M.Capus, R.M.German. Princeton : MPIF, 1992, p. 1

- [10] Seyrkammer, J., Blaimschein, F., Delarche, C., Pourprix, Y. In: Adv. Powder Metall.& Partic. Mat. 1992, vol. 5. Compiled by J.M.Capus, R.M.German. Princeton : MPIF, 1992, p. 141
- [11] Wantanabe, T., Yamada, K.: Int. J. Powder Metall., vol. 4, 1968, no. 3, p. 37
- [12] Kehl, W., Fischmeister, HF.: Powder Metall., vol. 23, 1980, no. 3, p. 113
- [13] Danninger, H., Neubing, HC., Gradl, J. In: Proc. PM'98 Powder Metall. World Congress & Exhib., Granada, vol. 5. Ed. V.Arnhold, A.Romero. Shrewsbury : EPMA, 1998, p. 272
- [14] Savitskij, AP.: Liquid Phase Sintering of the Systems with Interacting Components. Tomsk : Russian Academy of Sciences - Siberian Branch, 1993. 142 p.
- [15] Schaffer, GB., Huo, SH. In: Adv. Powder Metall. & Partic. Mater. 1996, vol. 4. Ed. T.M.Cadle, K.S.Narasimhan. Princeton : MPIF, 1996, p. 14.27
- [16] Sercombe, TB., Schaffer, GB. In: Powder Metall. & Partic. Mater. 1997, vol. 1. Ed. R.A.McKotch, R.Webb. Princeton : MPIF, 1997, p. 10.21
- [17] Schaffer, GB., Sercombe, TB., Lumley, RN.: Mater. Chemistry and Physics, vol. 67, 2001, p. 85
- [18] Neubing, HC., Danninger, H. In: Pulvermetallurgie in Wissenschaft und Praxis, vol. 14. Ed. H.Kolaska. Düsseldorf : VDI, 1998, p. 266
- [19] Mühlburger, M., Paschen, P.: Z.Metallkunde, vol. 84, 1993, no. 5, p. 346
- [20] Kehl, W.: Flüssigphasensintern von Preßlingen aus Al-Cu-Pulvern. Ph.D. Thesis. Leoben : Montanuniv., 1981
- [21] ASM Handbook, vol. 3. Materials Park : ASM, 1992, p. 3.10
- [22] Danninger, H.: Powder Metall. Int., vol. 24, 1992, no. 2, p. 73