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MODERN TECHNOLOGICAL METHODS OF PRESSURE WELDING OF MAGNESIUM ALLOYS (REVIEW)

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ABSTRACT

Proceeding from the results of literature analysis, the good prospects for application of pressure welding methods, namely diffusion, ultrasonic, and roll bonding, when joining elements from magnesium alloys, are shown. Comparative evaluation of bonding modes was performed. The main directions for producing sound joints were determined. It is shown that the main ways to improve the bonding processes are application of the following approaches: monitoring the temperature-time parameters of bonding, application of interlayers in the form of separate layers from similar materials, based on copper, nickel, zinc or silver, or of eutectic mixtures, butt joint strengthening through application of nano- or finely-dispersed particles, intensification of plastic deformation through superposition of ultrasonic oscillations, increase of plastic deformation intensity, as well as application of heat treatment before and after bonding.

KEYWORDS: magnesium alloys, welded joints, microstructure, strength, diffusion bonding, ultrasonic bonding, roll bonding

INTRODUCTION

Magnesium is one of the most widely spread elements on Earth. By estimates of total quantity, it is in the fourth place after iron, oxygen and silicon and in the eighth place in terms of its content in the earth's crust. The main advantage of application of magnesium alloys is their low weight: magnesium has one of the lowest densities among the structural materials [1]. Its active application is slowed down because of low ductility and formability at room temperature, presence of basic texture, edge cracking at rolling and high electrochemical activity [2].

Development of new alloys and introduction of modern processing methods promotes wider acceptance of magnesium in automotive and aviation industry. Over the recent years, magnesium alloys became wider applied in manufacture by large car companies, including General Motors, Ford, Volkswagen and Toyota. Dashboard elements, gearboxes, steering components and radiator supports are made from this group of materials. Such aircraft elements as brackets, compressor upper case for air conditioning mounting, door fasteners, twisted-blade impeller, antenna supports, window frame, parts of turbofan engine gearbox, etc. are already made from magnesium alloys [1, 2].

As magnesium becomes ever wider applied in manufacture of diverse components, there is the need to develop joining methods, which would provide the appropriate properties of the produced structural elements.

It is known, however, that the traditional technologies of fusion welding of magnesium alloys cause considerable softening of the material in the joint zone, formation of cast coarse-crystalline weld structure and of characteristic defects, such as pores, microinclusions of oxide films and cracks, which are due to metal melting and crystallization in the zone of permanent joint formation [3, 4].

Solid-phase welding processes can be effective methods of improvement of weld reliability, making impossible the formation of defects characteristic for fusion welding.

Considering the above-said, the objective of this work was to perform analysis of the techniques applied in pressure welding of magnesium alloys. These include rolling, diffusion and ultrasonic bonding and their combinations.

Figure 1 gives the main technological measures, which are used in roll, ultrasonic and diffusion bonding. Friction stir welding is not considered in this paper in view of the specific features of its impact on the joint zone, which make it essentially different from the above processes.

We will consider the publications of recent years devoted to this issue [5–31].

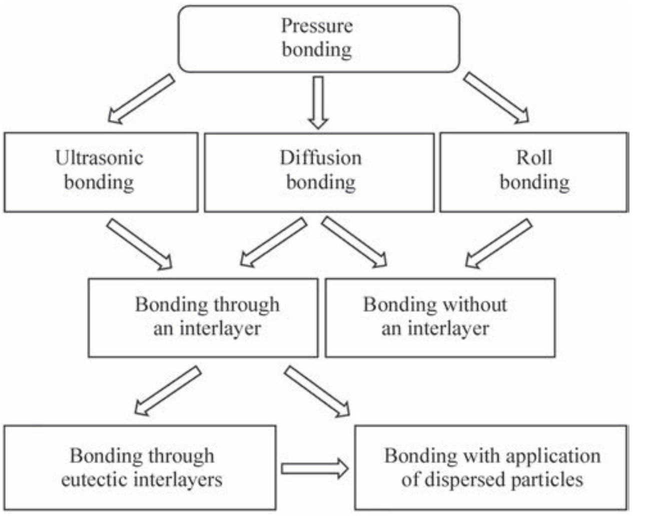

Figure 1. Main methods of pressure bonding of magnesium alloys

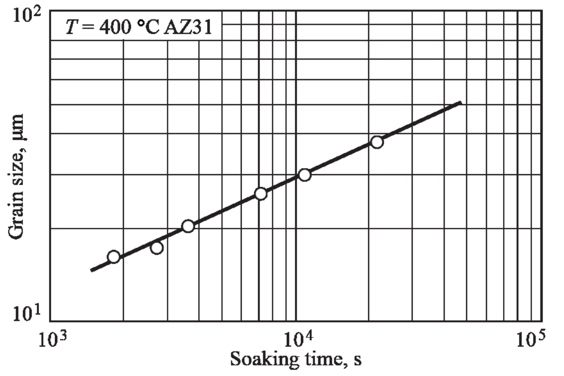

Figure 2. Interrelation between grain size and sample heating time at the temperature of 400 °C [6]

DIFFUSION BONDING

The main parameters in diffusion bonding (DB) are temperature, pressure and soaking time. In works [5–8] the influence of grain size and bonding parameters at sample heating in air on formation of the structure and mechanical properties of the joints were considered.

In work [5] when studying pure magnesium joints the following parameters were varied: temperature from 300 to 400 °C, pressure from 2 to 20 MPa, soaking time from 15 to 4320 min. Maximum values of shear strength of 95.7–109.4 MPa (88.8 % of base material strength) were achieved in the following mode: T = 400 °C, P = 20 MPa, t = 60 min. The joint line in such samples is not identified by optical microscopy. It is noted that the average grain size in the magnesium alloy was close to 85 μm in the initial condition.

When joining AZ31 alloy (Mg–3Al–1Zn, wt.%) [6], its preliminary annealing at 400 °C was performed for 30 min. Average grain size after this operation was 16.8 μm. Diffusion bonding was conducted in air at the temperature of 400 °C, pressure of 2–10 MPa, and soaking for 30–400 min. Here, superplasticity is observed in the alloy.

As shown by studies, conducted by the authors, increase of bonding time for samples from AZ31 magnesium alloy leads to increase of grain size (Figure 2).

During bonding, pores form between the contacting surfaces, the heating of which requires application of higher pressure or longer bonding time. Maximum values of shear strength of 74.2–80.5 MPa (85 % of that of the initial metal) were obtained in the following mode: T = 400 °C, P = 3 MPa, t = 180 min. The joint line in such samples is not revealed by optical microscopy.

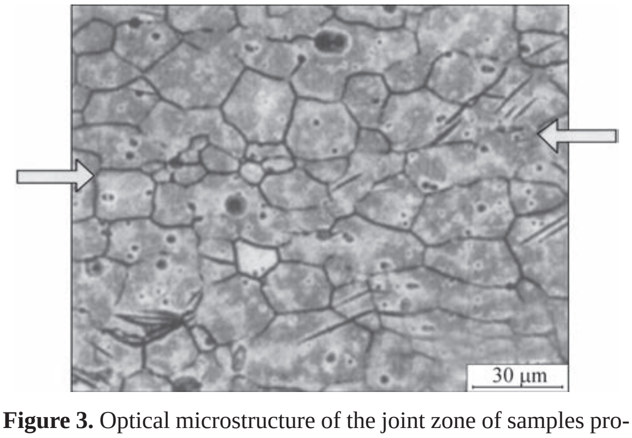

Figure 3. Optical microstructure of the joint zone of samples produced under pressure of 3 MPa for 180 min [6]

In work [7] studies of AZ31 alloy (Mg–3Al–1Zn, wt.%) were carried ont, when it demonstrated superplastic behaviour and at lowering of bonding temperature to 200–300 °C. Material was used after hot pressing and rolling. Average grain size was 8.5 μm. The authors note that the initial alloy grain size has an essential influence on the temperature, at which the superplastic properties are manifested, for instance for 16 μm grain size it is 300 °C, while for 8.5 μm it is 200 °C. Thus, such bonding temperature and pressure were selected, at which the alloy superplastic properties were manifested: 250–300 °C and 15–30 MPa, respectively. The process was conducted in air, and its duration was 60–120 min.

At bonding temperature of 250 °C, a clear contact line and pores are observed in the butt joint. Temperature rise allows lowering the defect rate in the joint zone. However, increase of soaking time up to 120 min leads to excess grain growth, and, consequently, to lowering of the produced joint strength. Maximum shear strength τ = 68.5 MPa (80.3 % of initial material strength) was achieved in the following mode: T = 300 °C, P = 20 MPa, t = 60 min.

The influence of grain size on formation of joints of AZ31 magnesium alloy at diffusion bonding was considered in [8]. Samples after annealing at 400 °C for 30 min with average grain size of 28 and 11 mm were used. Diffusion bonding was conducted in air at the temperature of 400 °C, pressure of 2–10 MPa, and soaking for 30–600 min.

It was shown that the grain size of the initial metal has an essential influence on the process of joint formation, and its coarsening requires application of a more stringent control of the bonding modes to produce strong joints. This is attributable to the dependence of metal superplastic behaviour on grain size. So, in the case of samples with grain size of 28 μm, the optimal mode is as follows: T = 400 °C, P = 5 MPa, t = 180 min, and for grain size of 11 μm it is: T = = 400 °C, P = 3 MPa, t = 120 min. Shear strength here is 74.5–81.1 MPa (90 % of that of the initial metal) and 81.4–85.1 MPa (92 %), respectively. The authors proposed a model of joint formation, which includes both the diffusion processes and plastic deformation of the subsurface layers. Recommendations are given for selection of bonding time at different pressure values, depending on base metal grain size.

In works [9–11] diffusion bonding of magnesium alloys was studied under the vacuum conditions.

In work [9] vacuum diffusion bonding (VDB) of extruded AZ31 magnesium alloy (Mg–2.5–3.5Al–0.5

–1.5Zn–0.2–0.5Mn — up to 0.1Si — up to 0.05Cu — up to 0.005Fe – up to 0.005Ni, wt.%) was performed. In order to obtain initial material with specified average grain size of 14.1 μm, preliminary isothermal annealing was conducted at 300 °C for 30 min. Bonding process parameters were assigned in the following ranges: temperature of 390–480 °C, pressure of 8 MPa, soaking time of 60–120 min, and degree of rarefaction in the chamber of 14 Pa.

Investigation results showed that bonding temperature and soaking time had a key influence on joint quality. At temperature below 420 °C, the diffusion processes in the butt joint are running slow, the contact line is clearly visible and the shear strength of such samples is low. At T = 450–480 °C partial disappearance of the contact line, essential grain growth and strength decrease take place. During soaking for 90 min, the diffusion processes are the most complete, sample strength grows, and further increase of bonding time leads to a slight decrease in strength, because of grain coarsening. The following mode was determined to be the optimal one: temperature of 420 °C, pressure of 8 MPa, and soaking for 90 min. Under these conditions, samples with maximum value of shear strength τ = 76.2 MPa were produced.

The authors of work [10] studied the features of joint formation at VDB of ZK60 magnesium alloy (Mg–6Zn–0.35Zr, wt.%). Average grain size was 11.9 μm. Bonding was conducted at the temperature of 300–400 °C, pressure of 1–20 MPa, soaking for 30–90 min, and vacuum of 0.27 Pa. Proceeding from the conducted mechanical investigations of the samples, the following mode was determined as the optimum one: temperature of 380 °C, pressure of 20 MPa, and soaking for 90 min. Shear strength of such joints is equal to 65.8 MPa (82 % of base metal strength). At lower values of mode parameters, the contact line and individual pores are visible in the butt joint. Exceeding the recommended values leads to excess growth of the grain, and, consequently, to a certain loss of strength. During bonding in the optimal mode, the average grain size increases from 11.9 to 21.2 μm. Up to certain limits, this process has a positive effect on sample quality, as the main mechanism of joint formation is atom diffusion and slipping of the initial grain boundaries, caused by grain growth.

In work [11] vacuum diffusion bonding of AZ61 magnesium alloy (Mg–5.8–7.2Al–0.4–1.5Zn–0.1–0.4 Mn — up to 0.3Si — up to 0.05Cu — up to 0.05Fe — up to 0.05Ni, wt.%) was performed. Preliminary recrystallization annealing at 300 °C for 30 min allowed reducing the average grain size from 20.48 to 14.29 μm. VDB was conducted at the temperature of 430–490 °C, pressure of 10 MPa, soaking time of 60–120 min. Degree of rarefaction is not given.

Metallographic investigations of samples, produced in the proposed modes, showed that a clear contact line remains in all the cases. However, its thickness decreases with increase of bonding temperature and pressure. At the same time, application of higher bonding temperature leads to excessive grain growth, and, consequently, to deterioration of the mechanical properties of the joints. Proceeding from experimental results, it was found that the maximum shear strength of 51.95 MPa (46.78 % of base metal strength) can be obtained in the following mode: T = 470 °C, P = 10 MPa, t = 90 min. Microhardness values of such samples are higher in the joint zone, than in the base metal (*HV* 82.48 against ~ *HV* 65) that is attributable to plastic deformation of the subsurface layers during bonding.

In work [12] it is proposed to perform heat treatment of the produced joints in order to improve their strength. The authors studied vacuum diffusion bonding of AZ91 alloy (Mg–9Al–1Zn, wt.). Recrystallization annealing was performed at 300 °C for 30 min with the purpose of grain refinement. Average grain size was 12.31 μm. Samples were bonded in vacuum of 18 Pa in the following mode: temperature of 430–490 °C, pressure of 10 MPa, and soaking time of 60–120 min. After bonding in the proposed modes, the contact line is clearly visible in all the samples, and considerable grain grown is found at a combination of high temperature with long-term soaking. Maximum shear strength of as-welded joints was equal to 64.7 MPa, respectively, and the following mode was determined to be the optimal one, respectively: T = 470 °C, P = 10 MPa, t = 90 min. Annealing of the joints was performed to improve their mechanical properties. Heat treatment was conducted at 320–380 °C for 60 to 300 min. An increase of their shear strength was achieved in all the modes, maximum values τ = 76.93 MPa were obtained at the temperature of 350 °C and soaking for 240 min. The joint line practically disappears here, the grains in the butt joint are fine, their size, however, gradually increasing with greater distance from the weld.

From the above works [5–12] on bonding magnesium alloys without using interlayers we can come to the conclusion that it is desirable to apply the bonding modes, not leading to considerable grain growth. Additional heat treatment allows somewhat increasing the joint strength, due to running of the recrystallization processes.

The possibility of application of interlayers of different chemical composition at diffusion bonding was studied in works [13–24].

In work [13] 50 μm thick copper foil was proposed as interlayer for VDB of AZ31B magnesium alloy (Mg–2.5–3.5Al–0.5–1.5Zn–0.2–0.5Mn–0.1Si–0.05 Cu–0.005Fe, wt.%). Bonding was performed at the temperature of 480 °C, pressure of 10 MPa, soaking for 30 min, and vacuum of 10^{-2} Pa.

Sound joints were produced, in the microstructure of which formation of three layers is observed (Figure 4).

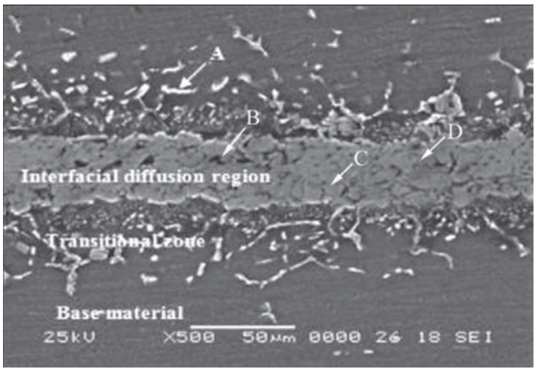

Figure 4. Microstructure of the joint zone of AZ31B alloy using a Cu interlayer, where A, B, C, and D are different areas of the diffusion zone [13]

The total thickness of the diffusion zone is 115 μm. The central layer of light-grey colour with black inclusions 35 μm thick is a mixture of Mg_2Cu (region D), $MgCu_2$ phases (region C) and Mg(Cu) solid solution (region B). The layers adjacent to the magnesium alloy (each 40 μm thick) are a dark-grey base with white, predominantly elongated inclusions (region A) and consist of Mg(Al,-Cu) solid solution and $Mg_{17}(Cu,Al)_{12}$ phase, distributed in it. Copper atoms demonstrate rather high activity in AZ31B alloy. On Mg/Cu interface copper diffuses along the grain boundaries into the base metal, forming a white compound. Joint formation is considered as a result of diffusion along the grain boundaries and dislocations, as well as solid-phase transformations. Microhardness distribution in the joint zone is of a steplike nature and gradually increases from the magnesium alloy to the central layer of the butt joint, where it is close to *HV* 100 that is by *HV* 50–60 higher than that in the base material.

The authors of work [14] applied a thinner layer of copper of thickness δ = 20 μm for bonding Mg–3Al–1Zn, wt.% alloy. Bonding was performed in argon atmosphere at the temperature of 530 °C, pressure of 0.7 MPa, and soaking for 5–120 min.

The authors divide the joint formation process into four stages: plastic deformation and diffusion in the solid state; dissolution of the interlayer and the base metal; isothermal crystallization and homogenizing (Figure 5, *a*–*c*). At the first stage (t = 5 min) drawing together of the contacting surfaces and their microplastic deformation take place. The second stage (t = 5–15 min) includes formation of the liquid phase as a result of a eutectic reaction that is accompanied by dissolution of the interlayer and part of the base metal. The joint zone here consists of solid solution of magnesium and $CuMg_2$ eutectic. At the stage of isothermal crystallization (t = 15–60 min), a gradual increase of the content of $CuMg_2$ phase occurs, as a result of intensive diffusion of copper into the base metal, that is accompanied by grain growth and appearance of a clear joint line in the center of the butt joint. The fourth stage is completion of the process of homogenizing of the joint zone, which is a solid solution of magnesium with $CuMg_2$ phase, distributed along the grain boundaries. The joint line losses its clear contours. Experimental results showed that the shear strength of the joints directly depends on bonding duration (Figure 5, *e*). With short bonding time (5–30 min) the mechanical properties of the joints are at a low level, because a too large quantity of brittle $CuMg_2$ phase remains in the butt joint, and at a too long bonding time (30–120 min) — because of intensive grain growth, combined with accumulation of $CuMg_2$ phase on grain boundaries. The following mode was determined to be the optimum one: T = 530 °C, P = 0.7 MPa, t = 30 min (Figure 5). Shear strength of the joints in this case is 70.2 MPa (85.2 % of base metal strength).

In work [15] bonding of AZ31 magnesium alloy was performed through an interlayer of copper and copper with tin. Copper 5 μm thick was applied on the sample surface by vacuum evaporation method. In part of the experiments tin foil 50 μm thick was additionally used. Bonding was conducted in argon at the temperature of 520 °C and pressure of 0.5 MPa, varying the soaking in the range of 10 50 min. Heating up to bonding temperature was conducted for approximately 2 min, and cooling to room temperature was performed together with the chamber.

It is shown that at application of the copper coating a diffusion zone 50–70 μm thick forms in the butt joint, its size decreasing with longer soaking time. It consists of solid solution of Mg and Cu_2Mg phase, and defects in the form of pores are present. Application of a combined interlayer of Cu coating and Sn foil allows producing a sound joint with a more uniform nature of element distribution in the butt joint and higher values of shear strength. The joint area consists of solid solution, enriched in Mg, and individual inclusions of Cu_2Mg phase. Sn layer does not take part in formation of intermetallic phases and freely diffuses in-depth of the magnesium alloy. Microhardness values in the butt joint are close to those of the base metal (68–70 VHN against ~ 58 VHN). Shear strength of the joints increases with longer soaking duration. However, after soaking for 30 min it almost does not change (τ = 64 MPa at t = 30 min and τ = = 67 MPa at t = 50 min).

Authors of work [16] used pure copper, as well as a combination of copper with nanoparticles of TiO_2, Al_2O_3 or SiC, as an interlayer in bonding AZ31 magnesium alloy (Mg–2.5–3.5Al–0.7–1.3Zn–0.3Si–0.2Mn –0.05Cu–0.005Fe–0.005Ni, wt.%). A galvanic coating from pure copper 20 μm thick was applied on the sample surface after cleaning. For additional deposition of the nanoparticles they were added to the electrolyte solution in the quantity of 10 g/l. TiO_2 and SiC nanoparticles of 20 nm size, and Al_2O_3 nanoparticles of 20 and 50 nm size were used. Bonding was conducted in the atmosphere of argon at the temperature of 525 °C, pressure of 1 MPa, and soaking for 60–120 min.

It is shown that increase of bonding time in all the cases promotes a more complete running of isother-

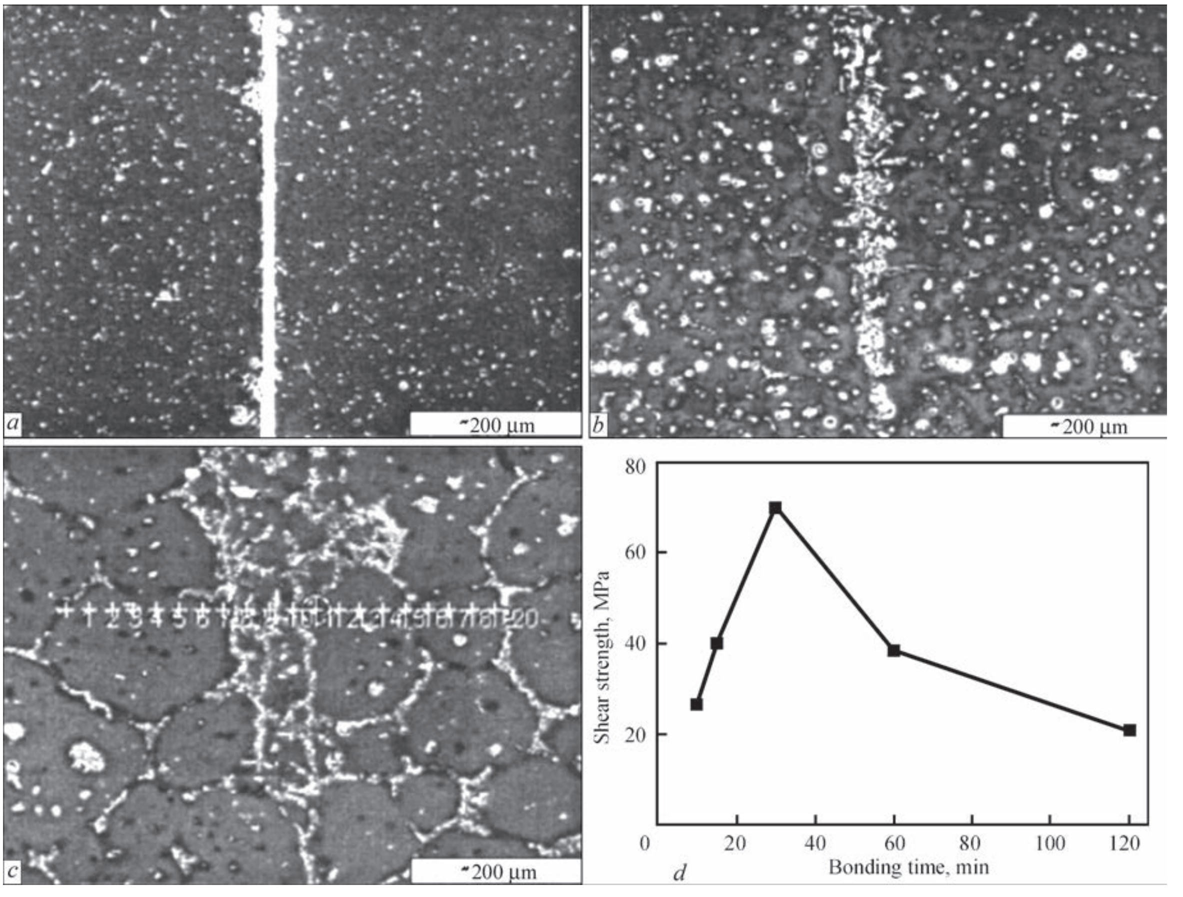

Figure 5. Microstructure of joints produced at 530 °C at bonding time of 5 min (*a*), 30 (*b*), 120 (*c*) and graph of dependence of shear strength on bonding time (*d*) [14]

mal crystallization process and allows producing a more homogeneous joint. When pure copper coating is used, a continuous interlayer of AlCuMg phase remains in the joint zone, along which microcracks are observed. At application of a copper interlayer with SiC particles a considerable quantity of Mg_2Cu phase forms in the central part of the butt joint, the presence of which is the cause for brittle fracture of such samples. In the case of an interlayer of copper with Al_2O_3 it was shown that increase of nanoparticle size from 20 to 50 nm leads to decrease of strength values. Here, a mixture of Mg_2Cu, AlCuMg phases and intermetallic phases of Al–Cu–Mg system, enriched in magnesium forms in the joint zone, and individual microcracks are observed. The highest shear strength of 31.66 MPa (98 % of that of the base metal) was found in samples bonded through an interlayer of copper with TiO_2 in the following mode: $T = 525$ °C, $P = 1$ MPa, $t = 120$ min. An accumulation of AlCuMg phase particles is observed in the joint zone (Figure 6). Uniform distribution of nanoparticles in the interlayer in this case inhibits crack initiation and propagation.

The authors of work [17] used aluminium foil 9 and 14 μm thick as an interlayer for bonding AZ31magnesium alloy (Mg–2.54Al–0.71Zn–0.09Si–0.03Mn–0.03Cr –0.03V–0.01Ti, wt.%). Bonding was performed at the temperature of 440–450 °C, pressure of 2 MPa, soaking for 45–120 min, and vacuum of 2.7 Pa.

The joint zone consists of three layers, and the aluminium content gradually decreases from the butt joint center to the base metal. At other conditions being equal, the joints, obtained through 9 μm foil have higher strength, that is attributable to a more complete running of the process of isothermal crystallization and formation of a smaller volume of $Al_{12}Mg_{17}$. At more than 75 min bonding duration, intensive grain growth occurs, leading to strength drop. The joint zone microhardness is up to three times higher than that in the base metal, while greater foil thickness leads to a certain increase of butt joint hardness. The following conditions were determined to be the optimal ones for joint formation: $T = 440$ °C, $P = 2$ MPa, $t = 75$ min and intermediate layer thickness of 9 μm. The shear strength was ~ 35 MPa.

The authors of work [18] considered the possibility of using silver foil 100 μm thick as an interlayer at vacuum diffusion bonding of AZ91 magnesium alloy (Mg–9Al–1Zn, wt.%). Bonding was conducted at the temperature of 480 °C, pressure of 1 MPa, and soaking for 30–120 min in the vacuum of $2\cdot10^{-3}$ Pa.

Sound joints were produced. During the bonding process, melting of the material on the boundary of AZ91 alloy/Ag foil and active diffusion of atoms between base material and the interlayer take place with silver gradually dissolving in the magnesium alloy matrix. The shear

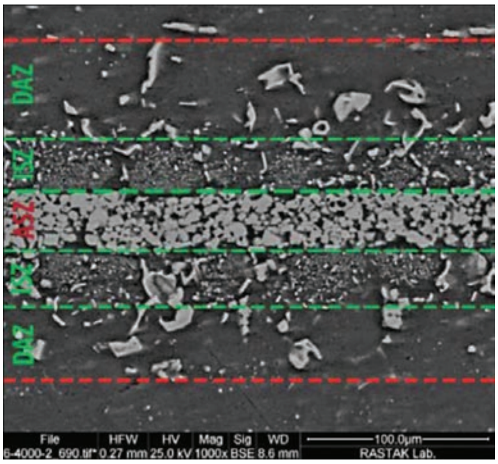

Figure 6. Microstructure of AZ31 alloy joint produced with application of a copper interlayer with TiO_2 nanoparticles [16]

strength of the joints is almost independent on bonding duration and it is in the range of 65–70 MPa (against ~120 MPa for base metal after annealing). The microstructure and chemical composition of the joint zone of samples produced at different soaking are similar.

In work [19] bonding of AZ31 magnesium alloy (Mg–2.99Al–0.96Zn, wt.%) was conducted through an interlayer of silver 50 μm thick. Bonding was performed at the temperature of 480–500 °C and soaking for 60 min.

In samples produced at T = 480–490 °C a eutectic layer is observed in the central part of the weld and diffusion layers are found at the boundary with base metal. In bonding at 500 °C just a thin eutectic layer forms in the joint zone. The composition of the found phases also changes with temperature rise. At 480 °C, a mixture of $AgMg_4$ intermetallic and $AgMg_3$ eutectic of the total width of 40 μm is formed in the butt joint. At 490° C a joint zone of approximately 100 μm width is formed, where the diffusion layers consist predominantly from a solid solution, enriched in Mg, and the central region consists of $AgMg_3$ eutectic. At 500 °C the interlayer with Ag almost disappears as a result of diffusion into the base metal, leaving a thin eutectic layer (~ 6 μm). It is noted that increase of bonding temperature also has a positive influence on microhardness of the joint zone, its value dropping from *HV*0.05–170 (*T* 480 °C) to *HV*0.05–80 (T = 500 °C). Shear strength is just 39 MPa at optimum bonding temperature that is two times lower than in the base metal.

The authors of [20] used a nickel interlayer at VDB of AZ31 magnesium alloy (Mg–3Al–1Zn, wt.%). The bonding process was conducted in the following mode: temperature of 515 °C, pressure of 0.16 MPa, soaking for 5–120 min, and vacuum of 0.053 Pa.

After bonding three diffusion layers, located symmetrically relative to the butt joint center, can be singled out in the joint zone. The central layer, which consists of a eutectic, gradually decreases and disappears with increase of soaking time to 60 min that is indicative of a complete isothermal crystallization of the weld. Directly adjacent to it are intermetallic layers, the thickness of which is increased with longer bonding time. At soaking in the range of 20–30 min the layers adjacent to the base metal, first become wider, and then decrease in size. It is attributable to gradual homogenizing of the joint zone, as interdiffusion of the interlayer and base metal elements takes place. Increase of bonding time to 120 min leads to strength drop that was related to formation and segregation of brittle intermetallics of Mg–Ni system. The following mode was determined as the optimum one: T = 515 °C, P = 0.16 MPa, t = 60 min. Under these conditions a joints with maximum shear strength of 36 MPa were produced, the microhardness being 179 VHN that is three times higher than in the magnesium alloy (55 VHN).

In work [21] AZ31 magnesium alloy (Mg–3Al–1Zn, wt.%) was bonded through a nickel coating. Nickel of 5 μm thickness was deposited by vacuum evaporation. Bonding was performed in the following mode: temperature of 520 °C, pressure of 8 MPa, soaking for 5–60 min, and vacuum of $1.33 \cdot 10^{-3}$ Pa.

In the first minutes of joint formation regions enriched in nickel are still observed in the butt joint. Increase of bonding duration up to 30 min promotes Ni diffusion and more complete passage of isothermal crystallization process. Maximum values of shear strength τ = 46.2 MPa were obtained at 20 min soaking. At greater duration of the process, intensive grain growth is observed in the butt joint alongside formation of Mg_2Ni brittle phase, leading to strength drop in such joints.

The authors of [22] considered the possibility of applying SiC particles as a reinforcing element of the butt joint in vacuum diffusion bonding of AZ91 magnesium alloy (Mg–8.25Al–0.63Zn–0.22Mn–0.035 Si–0.014Fe–0.03Cu–0.002Be, wt.%). Bonding was conducted at the temperature of 420 °C, pressure of 10 MPa, with 60 min soaking. SiC particles of 10–45 μm size in the ratio of 2–6 % were placed between the magnesium alloy surfaces.

It is shown that application of reinforcing particles allows increasing the shear strength almost two times, compared to bonding without interlayers (109 MPa against 63 MPa). The optimum size and concentration were determined to be 10–25 μm and 4 %, respectively. A too large volume of the particles leads to slowing down of the diffusion processes and SiC accumulation in certain regions of the butt joint that leads to lowering of ductility of such joints and their fracture at lower strength values.

It follows from the above that in bonding through eutectic interlayers it is necessary to control the process of brittle phase formation through reduction of bonding time, as well as limited formation of the liquid phase that is achieved by using thinner foils/coatings. Application of reinforcing particles of an optimum size and concentration can significantly increase the strength of the produced joints.

In works [23, 24] the influence of plastic deformation on welded joint formation was studied.

In work [23] bonding of AZ31 magnesium alloy ((Mg–3.04Al–0.99Zn–0.29Mn–0.02Si–0.01Fe–0.0Cu, wt.%)) was performed in two stages: hot deformation and annealing. Compression was performed at the temperature of 350 °C, strain rate of 0.1 s^{-1}, and degree of deformation of 10–60 %. The samples were heated to the specified temperature and compressed up to specified upsetting, leaving them under pressure for 3 min more. Then they were welded into vacuum tubes, and soaked for 400 °C at 60–720 min.

Increase of the degree of deformation to 40–60 % promotes passage of dynamic recrystallization of grains in the butt joint, and initial line of adhesion takes a wavelike shape. Growth of new grains at the interface and their gradual growth into adjacent deformation regions lead to migration between the grains and partial disappearance of the interface (Figure 7).

Annealing for 60 min leads to a certain growth of the recrystallized grains. Increase of soaking time to 120–240 min promotes formation of a more ramified contour of the grain boundaries, their size increasing only slightly. The interface disappeared completely after soaking for 480 min. Further increase of annealing duration leads to undesirable coarsening of the grain that has a negative influence on the mechanical properties of such joints. Maximum tensile strength of 164.7 MPa was found in samples, which were joined with 60 % degree of upsetting and annealed for 480 min that is by 9% higher than in samples soaked for 60 min.

The authors of work [24] studied the features of formation of joints of WE43 magnesium alloy (Mg–4Y–3Nd–0.5Zr, wt.%). Samples were heated up to 450–525 °C and compressed, setting the strain rate of 0.001 — 1 s^{-1} and degree of strain of 5–50 %. After heating to the specified temperature, they were compressed and quenched in water.

In the joint zone a rapid dynamic grain growth takes place under the conditions of low strain rate of 0.001 s^{-1} and high bonding temperature of 525 °C, leading to strength lowering (σ_t = 114.43 MPa). In case of conducting the process at the temperature of 450 °C or high strain rate, grain refinement in the joint zone is controlled by dynamic recrystallization.

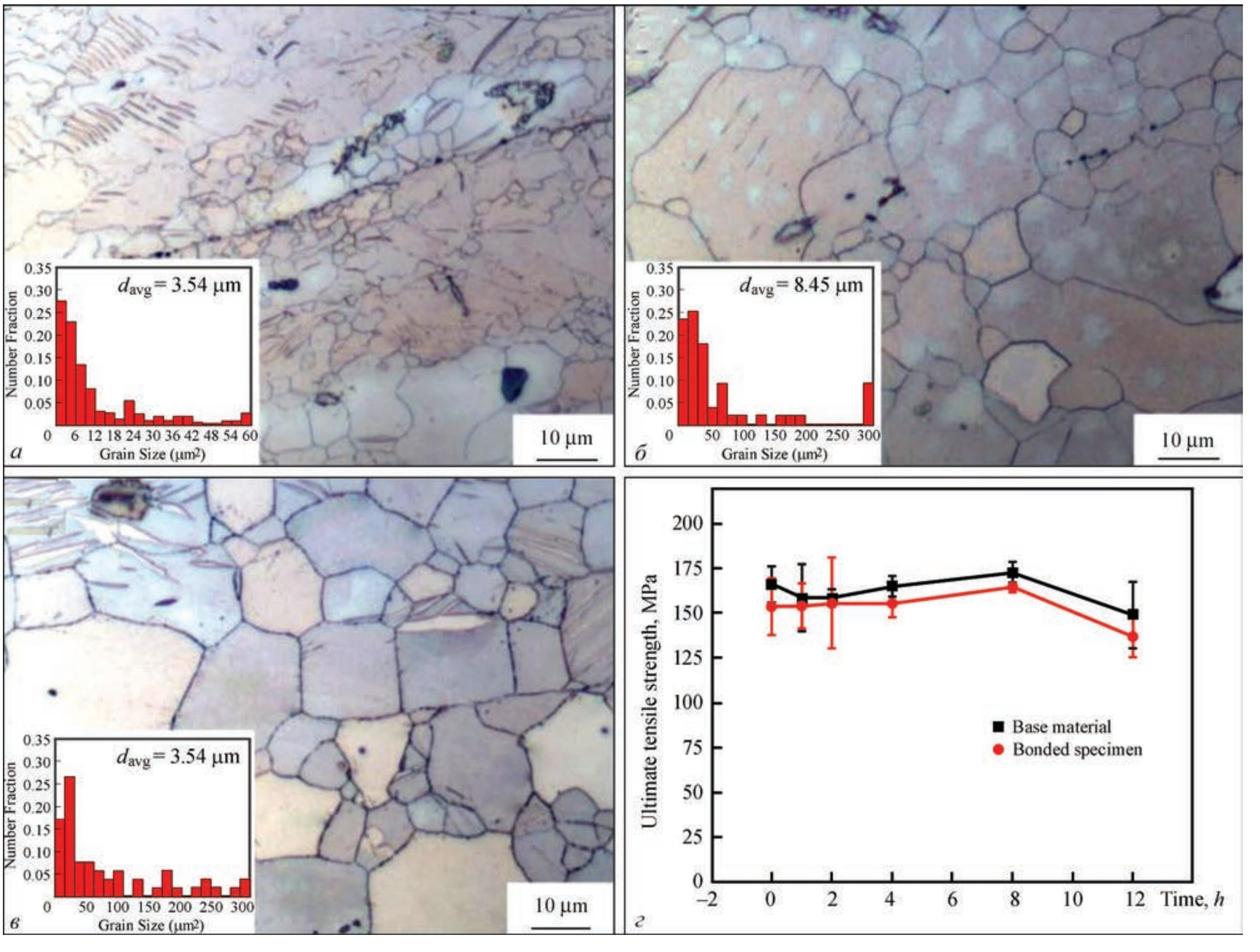

Figure 7. Microstructure of the joint zone and distribution of grain size in samples produced at the temperature of 350 °C, at the degree of deformation of 0.6 with subsequent heat treatment at 400 °C for: *a* — 60; *b* — 240; *c* — 480 min. Graph of the influence of heat treatment time on shear strength of the joints (*d*) [23]

It is shown that bonding performance at a high strain rate of 1 s^{-1} leads to a considerable drop of ductility, compared to the initial alloy (3.56–5.51 % against 12.17 %). The highest values of rupture strength of 158.60 MPa were obtained for samples joined at the temperature of 450 °C and strain rate of 0.001 s^{-1}, with ductility being 6.99 %.

ULTRASONIC BONDING

In the works on ultrasonic bonding of magnesium alloys [25–28] oscillations are additionally used, alongside parameters characteristic for diffusion bonding (heating temperature, pressure) that allows essentially shortening the bonding time.

In work [25] bonding of ME20M magnesium alloy (Mg–1.3–2.2Mn–0.15–0.35Ce–0.3Zn, wt.%) was conducted through an interlayer of pure Zn 50 μm thick. Samples from ME20M alloy were welded in the inductor in air at the temperature of 370 °C, pressure of 0.15 MPa, and soaking up to 2 min. Ultrasound of 200 W power at the frequency of 20 kHz was used as an additional source of intensification of joint formation.

It was not possible to produce joints in the proposed mode without ultrasonic oscillations. Application of ultrasound promotes breaking up of the oxide film at contact surfaces, accelerates the eutectic reaction, intensifies Mg and Zn diffusion in the eutectic liquid and accelerates the process of isothermal crystallization. Already after one second of ultrasound action, the zinc interlayer transforms into a two-phase gray layer, the main part of which consists of $Mg_{51}Zn_{20}$ with acicular MgZn inclusions. Increase of bonding time (Figure 8) causes a gradual decrease of the thickness of the central eutectic layer and its disappearance as a result of completion of the process of isothermal crystallization with formation of Mg(Zn) solid solution with individual inclusions of MgZn phase. The highest average shear strength of the joints of 106.4 MPa was obtained for samples, where bonding time was 2 min.

In publication [26] foil of N_{26} alloy (Cu–38Zn) 20 μm thick was used as an interlayer. Bonding was conducted at the temperature of 460 °C, pressure of 0.15 MPa, process duration of 0.05–1.5 min, ultrasound frequency of 20 kHz and power of 500 W.

It is shown that increase of bonding time promotes a gradual dissolution of the interlayer with formation of $CuMg_2$ and CuMgZn phases, which, in their turn, disappear as a result of formation of a solid solution in the butt joint. The process of isothermal crystallization is fully completed after 1.5 min. The oxide films are fragmented, partially pressed out of the butt joint with the liquid phase, and their remains are arranged in two strings along the base metal. Samples, produced at bonding time of 1.5 min, had shear strength at the level of base metal (105 MPa).

The authors of [27] studied the possibility of ultrasound application in bonding AZ31B magnesium alloy. Zn foil 0.5 mm thick was used as an interlayer. Oscillation frequency was 20 kHz. Bonding was conducted in air at the temperature of 360–380 °C, pressure of 0.36 MPa, and soaking of 0.02 min.

It was shown that at bonding temperature of 360–380 °C a layer of $MgZn_2$ intermetallic with a string of pores in the middle, eutectic layers on the boundary of contact with the base metal and thin MgZn inclusions on the interface between $MgZn_2$ and eutectic form in the center of the butt joint. Here, temperature rise promotes thinning of the central layer. At 380 °C the zinc foil reacts completely with magnesium, forming a eutectic structure with inclusions of magnesium-based solid solution. Maximum shear strength of 42 MPa was demonstrated by samples produced at the temperature of 380 °C.

The work shows a diagram of joint formation through a liquid interlayer (Figure 9). Dense oxide layers were present on the surfaces of magnesium alloy and Zn before bonding. After heating up to bonding temperature and switching on the ultrasonic oscillations, the oxides start breaking up, and a liquid phase forms in their place. Active diffusion of Mg and Zn promotes a fast formation of a large quantity of liquid, breaking up of oxide remains and their pressing out into flash. Mg atoms diffuse into Zn interlayer, forming $MgZn_2$ and MgZn in the solid state. The liquid phase forms a eutectic structure, at the boundary with the base metal.

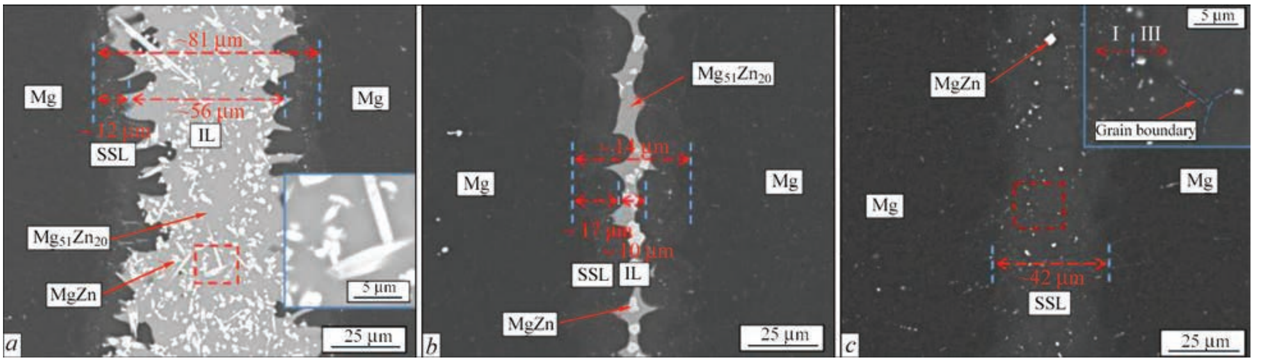

Figure 8. Microstructure of the joint of ME20M alloy, using Zn interlayer and ultrasonic oscillations at process time of: *a* — 0.02; *b* — 0.5; *c* — 2 min [25]

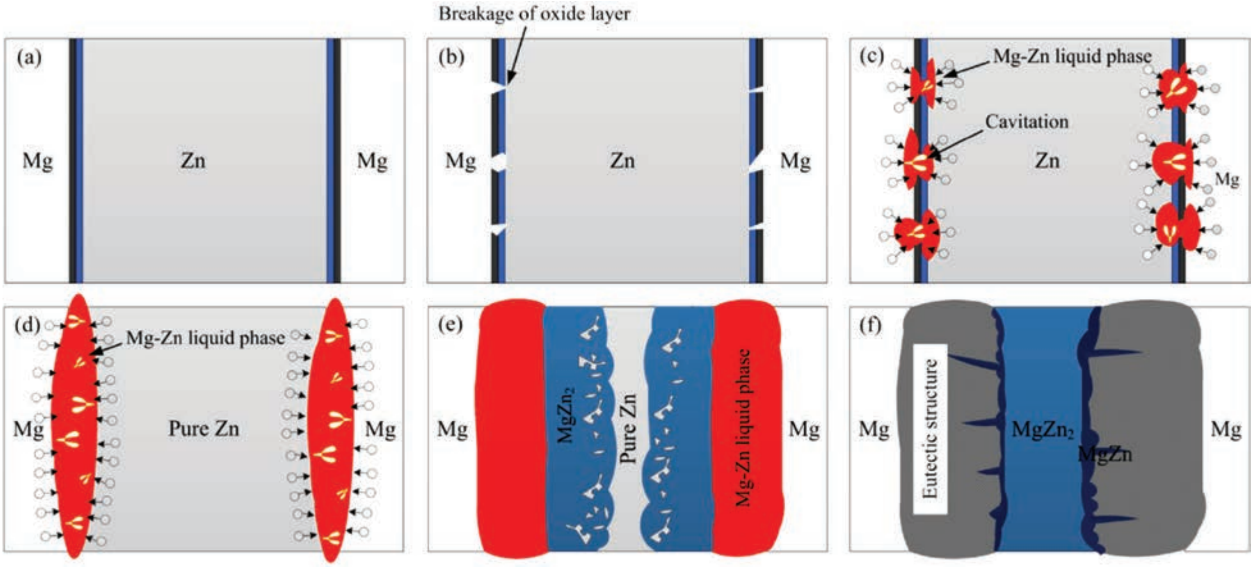

Figure 9. Scheme of the mechanism of joint formation through a liquid interlayer [27]

In work [28] the influence of pulse power on the quality of joints of AZ31B magnesium alloy was studied. Zinc foil 0.5 mm thick was used as an interlayer. Bonding was conducted in air at the temperature of 360 °C and pressure of up to 0.4 MPa. Working frequency was 20 KHz with the power of 333–1000 W.

Investigations showed that the oxide is completely removed from the magnesium alloy surface already after one second of ultrasound action. A wide joint zone (~1.2 mm) forms in the butt, which consists of a layer of $MgZn_2$ with inclusions of Zn in the central part and of Mg/Zn eutectoid structures on the boundary with the base metal, and MgZn inclusions are observed between both the layers. Increase of bonding time promotes a reduction of the size of the central zone, from which individual longitudinal inclusions remains already after 3 s. In the joints produced at higher power of ultrasound, smaller residual porosity and greater width of the butt joint with thicker eutectoid layers are observed. At sonotrode pressure of 0.4 MPa a joint zone 120 μm wide, completely consisting of the eutectoid phase, forms in the butt joint. Investigations of mechanical properties showed that joints, which completely consisted of the eutectoid structure, had the highest shear strength of 40 MPa.

Application of ultrasound at magnesium alloy bonding allows reducing the duration of the process of joint formation several times, compared to diffusion bonding, due to intensification of the diffusion processes and rapid destruction of oxide film on contact surfaces. However, application of this method is limited by the ability to weld small-sized samples.

ROLL BONDING

In works [29–31] on roll bonding the degree of sample strain was considerably increased from 32 % [29] to 50 % [30, 31], unlike diffusion bonding, where the degree of strain is equal to 2–5 %.

The authors of work [29] studied the influence of process temperature at accumulative rolling of AZ31 alloy. Sheets of 1 mm thickness were used as initial metal, which were stacked into packs of three and heated in the furnace up to 350–450 °C for 10 min. Rolling was performed with 32 % strain and 14 m/min rate.

It was shown that at preheating of sample assembly up to 350 °C, dynamic recrystallization starts taking place at the grain boundary. At 400 °C shear bands start forming, which consisted of dynamically recrystallized grains. At 450 °C a shear band of irreg-

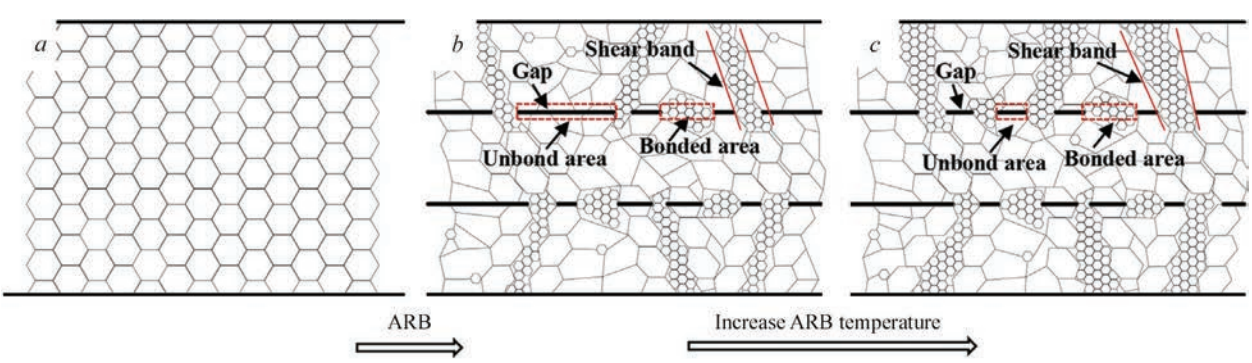

Figure 10. Schematic illustration of evolution of the microstructure and mechanism of formation of AZ31 alloy joint at different process temperature

Parameters of pressure bonding of magnesium alloys

Material	Bonding process	Interlayer	Bonding parameters			Joint strength, MPa	Source
			T, °C	*P*, MPa	*t*, min		
Pure Mg	DW	–	400	20	60	τ = 95.7–109.4	[5]
AZ31	DW	–	400	3	180	τ = 74.2–80.5	[6]
AZ31	DW	–	300	20	60	τ = 68.5	[7]
AZ31	DW	–	400	5	180	τ = 74.5–81.1	[8]
			400	3	120	τ = 81.4–85.1	
AZ31	DW	–	420	8	90	τ = 76.2	[9]
ZK60	DW	–	380	20	90	τ = 65.8	[10]
AZ61	DW	–	470	10	90	τ = 51.95	[11]
AZ91	DW	–	470	10	90	τ = 76.93[1]	[12]
AZ31B	DW	Copper foil	480	10	30	–	[13]
Mg–3Al– 1Zn	DW	Cu foil	370	0.7	30	τ = 70.2	[14]
AZ31	DW	Cu coating + Sn foil	520	0.5	50	τ = 67	[15]
AZ31	DW	Cu coating + TiO_2 nanoparticles	525	1	120	τ = 31.66	[16]
AZ31	DW	Al foil	440	2	75	τ = 35	[17]
AZ91	DW	Ag foil	480	1	30–120	τ = 65–70	[18]
AZ31	DW	Ag foil	500	–	60	τ = 39	[19]
AZ31	DW	Ni	515	0.16	60	τ = 36	[20]
AZ31	DW	Ni coating	520	8	20	τ = 46.2	[21]
AZ-91	DW	SiC particles	420	10	60	τ = 109	[22]
AZ31	DW	–	350	–[3]	3	σ_t = 164.7[1]	[23]
WE43	DW	–	450	–[3]	–	σ_t = 158.60	[24]
ME20M	USW	Zn foil	370	0.15	2	τ = 106.4[2]	[25]
ME20M	USW	Cu–38Zn foil	460	0.15	1.5	τ = 105[2]	[26]
AZ31B	USW	Zn foil	380	0.36	0.02	τ = 42[2]	[27]
AZ31B	USW	Zn foil	360	0.4	–	τ = 40[2]	[28]
AZ31	RB	–	450	–	–	σ_t = 295.4	[29]
LZ91	RB	–	150	–	–	σ_t = 290.2	[30]
AZ31	RB	–	400	–	–	–	[31]

Notes. [1]Strength values after joint heat treatment. [2]Application of ultrasound as an additional source of intensification of joint formation process. [3]Compression in bonding with the specified level of upsetting.

ular shape forms, the width of which decreases from the surface inwards. Increase of sample temperature before rolling intensifies the dynamic recrystallization in the butt joint and improves the joint quality. At a high temperature, grain boundaries shift easily, and the recrystallized grains come into contact and coalesce. The features of joint formation are shown schematically in Figure 10, depending on the preheating temperature. On the whole, the rolling process significantly lowers the material elongation value and almost does not change the yield limit and rupture strength. An essential lowering of ductility is attributable to cracking during rolling. The optimal preheating temperature was 450 °C. At this temperature the yield limit was 222.8 MPa, ultimate strength was 295.4 MPa and elongation was 6.0 %.

In work [30] LZ91 magnesium alloy (Mg–6Li–1Zn) was subjected to accumulative rolling. Plates 100 mm long and 0.65 mm thick were cleaned from one side, stacked into packs of two, heated up to 150 or 400 °C in air and rolled with the degree of strain of 50 % at roll temperature of 135 °C. Then, the thus obtained two-layer sheets were again stacked into a pack of two and the process was repeated up to 5 times.

In the initial state an elongated fibrous structure is found in LZ91 alloy, which consists of two main α and β phases. Pack rolling promotes grain refinement and formation of a more homogeneous structure. At preheating temperature of 400 °C and five rolling cycles the fibrous structure of Mg-rich α-phase becomes thinner and discontinuous. Samples preheated up to 150 °C and rolled five times had the maximum value of ultimate strength of 290.2 MPa. Application of a higher preheating temperature leads to a certain drop of hardness and strength values, as a result of passage of the process of recovery and recrystallization.

The authors of [31] studied the deformational behaviour at accumulative rolling of AZ31 magnesium alloy (Mg–3.2Al–1.3Zn–0.4Mn–0.06Ce–0.03Fe–0.01 Si, wt.%). Sheets of 2 mm thickness were stacked into packs of two and rolled with the degree of strain of 50 % and rate of 24 mm/min. Preheating temperature was 350–400 °C.

Joints could only be produced at preheating up to 400 °C. After the first pass a line of contact with dispersed particles of the oxide is clearly visible in the butt joint. Al_8Mn_5 phase is rather uniformly distributed in the base metal and joint zone. Cracks form at the edges. After the second pass, the adhesion strength was improved, and no delaminations were observed on the interface. Rolling essentially influences the grain size: already after the first pass, their average size drops by an order of magnitude from 100 to 10 μm, and after the second one — to 7.6 μm, here coarse grains practically disappear. Grain refinement occurs due to rotational dynamic recrystallization. It is emphasized that increase of yield limit and ultimate strength values occurs even after one rolling pass. Certain heterogeneity of the mechanical properties is noted that is related to the texture formed during rolling: they are somewhat higher in the rolling direction, than across it. This anisotropy decreases with increase of the number of passes. While deformation of samples cut out in the rolling direction is mainly performed by dislocation slipping, mechanical twinning has an important role during deformation of samples cut out across the rolling direction.

Results of investigations presented in works [29–31] for roll bonding, lead to the conclusion that increase of plastic deformation rate leads to an essential grain refinement and improvement of joint strength due to running of dynamic recrystallization processes.

Proceeding from the currently available technologies of joining magnesium alloys (Table), we can conclude that there are no well-established technologies at their bonding to produce sound joints. Modes which differ essentially from each other are proposed even for bonding similar alloys. On the whole, it can be summed up that it is desirable to conduct the process as lower bonding temperatures and/or bonding duration, and to allow for the initial metal grain size. Application of eutectic interlayers allows significantly shortening the time of joint formation.

CONCLUSIONS

Pressure bonding processes are extensively applied to produce magnesium alloy welded joints. The main ways to improve the bonding processes are application of the following approaches:

- control of temperature-time parameters of bonding to make grain growth impossible;
- application of interlayers in the form of individual layers of similar materials based on copper, nickel, zinc, aluminium or silver, or eutectic mixtures that allows, due to component interaction, accelerating destruction of the oxide film on the surface of samples being welded;
- butt joint strengthening due to application of nano- or fine particles;
- intensification of plastic deformation due to superposition of ultrasonic oscillations;
- increase of plastic deformation intensity;
- application of heat treatment before and after bonding.

Thus, in diffusion bonding of magnesium alloys the following can be considered promising: application of thin interlayers or coatings that should reduce the chemical heterogeneity in the butt joint and promote activation of the contact surfaces, strengthening of the joint zone due to addition of fine particles to the butt, or their formation during an increase of the rate and degree of plastic strain.

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CONFLICT OF INTEREST

The Authors declare no conflict of interest

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