https://doi.org/10.37434/tpwj2020.04.02

PECULIARITIES OF PRODUCING LAYERED METAL COMPOSITE MATERIALS ON ALUMINIUM BASE

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Analysis of publications devoted to producing layered composite materials on aluminium base was performed. The methods of joining thin-foil materials were studied, which allow producing layered joints with different number of intermediate layers. It is shown that the main welding methods, allowing production of joints with a layered structure, are rolling, ultrasonic, explosion and diffusion welding, and reaction sintering. Analysis of publications showed that the joining process can be conducted both in vacuum and in air. Joining foil from titanium and aluminium in the welding modes below the aluminium melting temperature (660 °C), allows producing joints without intermediate phase formation between the layers. In order to improve the strength of the produced composites, it is rational to apply during welding a technological operations in the form of current passing or postweld heat treatment of layered composite materials that provides increase of reactivity of the aluminium and titanium layers and formation of intermetallic phase as the reaction products. 31 Ref., 7 Figures.

Keywords: metal layered composite materials, joining methods, joint zone, intermetallic phase

When samples of new equipment are created, ever higher requirements are made of its weight and strength values that necessitate development of new materials which would satisfy this demand. Composite materials can be regarded as such.

Composite materials (CM) are those which have in their composition components insoluble or little soluble in each other, which differ essentially by their properties and which are separated by a clearly defined boundary in the material. CM properties mainly depend on physico-mechanical properties of the components and strength of the bond between them. A distinctive feature of CM is the fact that component advantages and not their disadvantages are manifested in them. At the same time, CM have properties, not found in the components which are included into their composition. In order to optimize the properties, components that differ markedly, but complement each other are selected [1].

CM properties depend on the shape and nature of filler distribution. By their geometrical shape, the fillers are divided into zero-dimensional, one-dimensional and two-dimensional (Figure 1). Zero-dimensional fillers have sizes of the same order in three dimensions (dispersion-strengthened CM). One-dimensional fillers consist of fibers of different intersections (fibrous CM), and two-dimensional fillers are plates, the length of which is greater than their thickness (layered CM) [2].

Layered composite materials (LCM) consist of rigidly connected metal or metal-containing alternating layers. These materials have a unique laminated structure with clear-cut interfaces and they can consist of layers of a broad range of materials. Depending on their thickness, the composite layers can be classified as sheets or plates of 1–10 mm and greater thickness, foils of 0.05–1.0 mm thickness and films of 0.001–0.05 mm thickness [3]. LCM can greatly improve a number of properties, including fracture toughness,

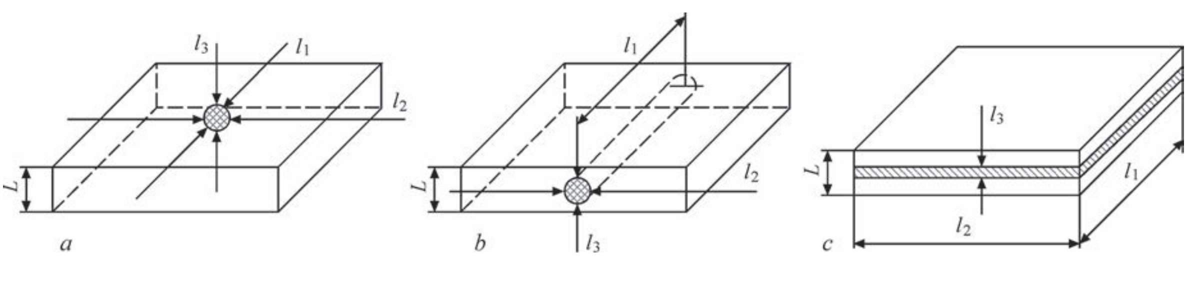

Figure 1. Forms of fillers: *a* — zero-dimensional; *b* – one-dimensional; *c* — two-dimensional: l_1; l_2; l_3 — filler dimensions; *L* — matrix thickness [2]

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fatigue characteristics, impact characteristics, wear, corrosion and damping ability, and provide higher ductility of brittle materials. Selection of component materials and their volume percent, layer thickness, and respective treatment, allows producing a material with specified properties [4].

LCM with a hard matrix and soft filler, are mainly applied as high-temperature materials, and those with a soft matrix and hard filler are used as heat-resistant materials. By functional characteristics LCM are divided into: corrosion-resistant, antifriction, electrical engineering, tool, wear-resistant, thermobimetals, bimetals for deep drawing and household products [3].

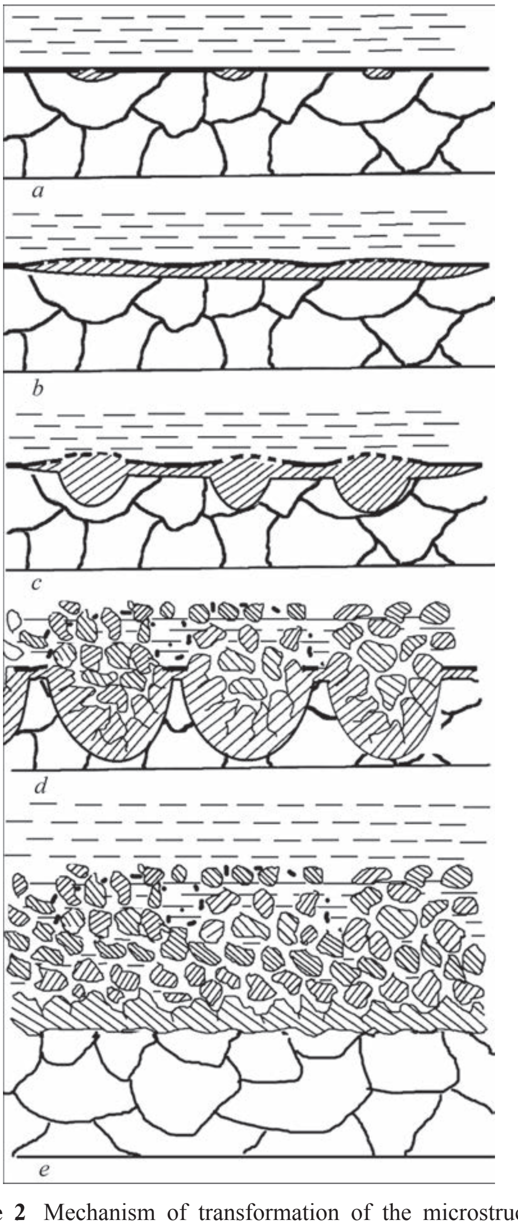


Figure 2 Mechanism of transformation of the microstructure of VT1-0/AD1 layered composite during heat treatment at temperatures above aluminium liquidus point: *a* — aluminide nuclei formation; *b* — transverse growth of thin interlayer sections; *c* — formation and acceleration of the growth of intermetallic phase nuclei in the zones of channels between oxide film fragments; *d* — intermetallic fragmentation; *e* – appearance of a continuous band of intermetallic fragments coated by a film of Al melt [5]

Considering that aluminium and titanium are some of the lightest metals, and are widely applied in different industries, the objective of this work was performing analysis of the methods to produce LCM based on aluminium and titanium.

The methods to produce LCM can be conditionally divided into those, where materials are formed in the solid or liquid state by welding methods, as well as through gaseous condition or solution by deposition methods. The following can be included into the first group: joining by melt impregnation techniques, sintering, rolling, ultrasonic, diffusion and explosion welding. The second group includes: thermal spraying, evaporation and vapour phase deposition in vacuum, methods of chemical deposition of metal from melts.

Let us consider producing layered composite materials by welding methods.

LCM manufacture by **melt impregnation method** implies contact of liquid filler metal with the solid metal of the matrix. The authors [5] proposed a mechanism of running of the processes of aluminium melt interaction with solid titanium (Figure 2), in keeping with which the aluminium melt is not directly contacting titanium initially, as Al_2O_3, TiO_2 (or TiO) interlayers are located between them. The process of titanium and aluminium atom diffusion through the interface in the presence of oxide films is slowed down. That is why it predominantly occurs along the ruptures of oxide films with appearance of local areas of oversaturated solid solution of Ti(Al) in titanium with further formation of nuclei of Al_3Ti intermetallics. This is followed by transverse growth of aluminides up to their coalescence with formation of a continuous interlayer along titanium.

Oxide film fragmentation promotes diffusion of aluminium atoms from the melt to the titanium surface, thus promoting formation and accelerated growth of intermetallic phase nuclei. Intermetallic formation in the volume limited by titanium which has reacted, leads to further breaking up of the oxide film, abrupt increase of internal stresses and breaking up of the formed aluminides into individual fragments, and their ousting from the reaction volume. This facilitates access of the aluminium melt practically to the reaction surface and promotes acceleration of the processes of localized growing of intermetallics right up to their closure and formation of a continuous strip of aluminide fragments, coated by a film of aluminium melt. Fragments of intermetallics that broke off, are inactive, and stay near the reaction surface, from which they are driven by new fragments that form constantly.

Work [6] is a study of the impact of liquid phase temperature on LCM formation. The materials used were commercial aluminium A7 and VT1-0 titanium of 40×40×0.5 mm size. The gap between the titanium layers was 0.125–0.7 mm. Titanium was first etched for 5–10 min at 20 °C in the following solution: 1 l of water, 20–30 ml H_2NO_3 and 30–40 ml HCl. Flux of KF–AlF_3 system was used for surface activation. Temperature of aluminium melt was set in the range of 670–950 °C. It was found that the wetting level practically does not change at increase of melt temperature and reaches the maximum value at 700–750 °C. The highest values of wetting level, as well as better mechanical properties in the entire studied temperature range were obtained for joints with 0.35 mm gap between the titanium layers ($\tau \approx$ 65–80 MPa). During LCM formation a two-phase zone is created between Al and Ti layers, which consists of aluminium (Al and up to 0.5 % Ti) and titanium (Al_3Ti) phases. Here, the quantity of the intermetallic phase increases with process temperature. At 800 °C a continuous interlayer of Al_3Ti forms between aluminium and titanium.

When LCM are manufactured by **plasma sintering** method, pulsed direct current is used as the heat source. Here, the material in the impact zone can be heated up to high temperatures, right up to plasma state.

In work [7], layered titanium-aluminium composite was produced by the method of plasma sintering of aluminium and titanium foil of 26 mm diameter, 100 and 200 μm thick, respectively. Total number of foils in the pack was 70 pcs. The process was conducted at the temperature of 830 °C, pressure of 5.7 MPa, and 1–5 min duration. Partial pressing out of molten aluminium from the pack, as well as plastic deformation of titanium lead to reduction of sample thickness by 10–30 %. Process performance results in formation of layered Ti−Al_3Ti intermetallic composite. Increase of soaking time up to 50 min, allows reducing porosity in titanium aluminide layers, as well as increasing their hardness from 2.9 up to 3.8 GPa.

In order to obtain the layered structure, the authors of [8] also used plasma sintering. Investigations were conducted with foil of commercial aluminium and titanium of 27 and 45 μm thickness, respectively, and 15 mm diameter. Foil was cleaned in an ultrasonic pool in methanol and assembled in a pack of 40 Al discs and 41 Ti one. Spark plasma sintering was conducted under vacuum of 10^{-3} Pa at direct current of 1.0–1.5 kA, voltage of 5–10 V; pulse frequency changed in the range of 30–40 kHz. The pack was first sintered at the temperature of 450 °C, pressure of 50 MPa, and soaking for 10 min. This was followed by annealing at 900 °C, for 30 min in an argon atmosphere. Final sintering was performed in the temperature range of 950–1200 °C at heating rate of up to 100 °C/min, pressure of 50 MPa, at soaking for 10 min. Preliminary sintering resulted in formation of a layered structure without any noticeable interaction of Al and Ti. After annealing the binary structure is transformed into the following interlayers: α-Ti, $AlTi_3$, AlTi, Al_2Ti and thick layer of porous Al_3Ti. Al melting, Kirkendale effect and difference in molar volumes between the reagents and products are indicated as the causes for formation. The second sintering cycle allows almost complete healing of pores. At process temperature of 950 °C the following layer composition forms: α-Ti, $AlTi_3$, AlTi, $Al_{11}Ti_5$ and Al_3Ti. Further temperature rise leads to reduction of the number of layers. So, α-Ti, $AlTi_3$, AlTi were obtained at 1050 °C, and a homogeneous mixture of $AlTi_3$ with AlTi was produced at 1200 °C. It was also established that the bending strength of layered samples produced at 950 °C is more than three times higher than that of samples sintered at 1050–1200 °C ($\sigma_{t\ bend}$ = 1400 MPa against $\sigma_{t\ bend}$ = = 389–446 MPa). This is accounted for by availability of α-Ti layer, which deformed plastically and prevented crack propagation in the direction normal to the composite layer plane, promoting crack deviation and branching. Such a microstructure absorbs a large amount of energy before breaking.

In work [9] reaction sintering of a foil pack from VT00 and A5E alloys (100×10×0.05 mm) was performed. The first and last layers were from titanium. Total thickness of the pack was 1 mm. Foils were first processed in 2 % solution of hydrofluoric acid. The pack was placed between ceramic plates and heated up to 630 °C in a heat-resistant box at the rate of 15 °C/min. Current was passed with a frequency of 40, 800 and 1600 Hz. The best results were obtained at application of unipolar pulses («−» was connected to Ti layers, «+» was connected to Al). It was found that at heating without current passage, complete transformation of the pack into Ti−Al_3Ti LCM occurs in 420–480 min. Additional passage of current reduces the time of layered composite formation to 300 min. During sintering, just Al_3Ti intermetallic forms. Its formation leads to appearance of pores and microcracks at the start of chemical transformation. Before completion of the sintering process, the discontinuities heal under the impact of pressure so that the final structure of layered composite is an alternation of residual Ti layers 30–35 μm thick and Al_3Ti layers with microhardness of 4.5–7.0 GPa.

The authors of [10] proposed a procedure of calculation of process parameters and scheme of layer transformation when producing titanium-aluminium LCM at direct current passage. Aluminium and titanium foils (90×10×0.05 mm) were assembled in a pack of the total thickness of 1 mm. The first and last layers were from Ti. Sintering was conducted at the temperature

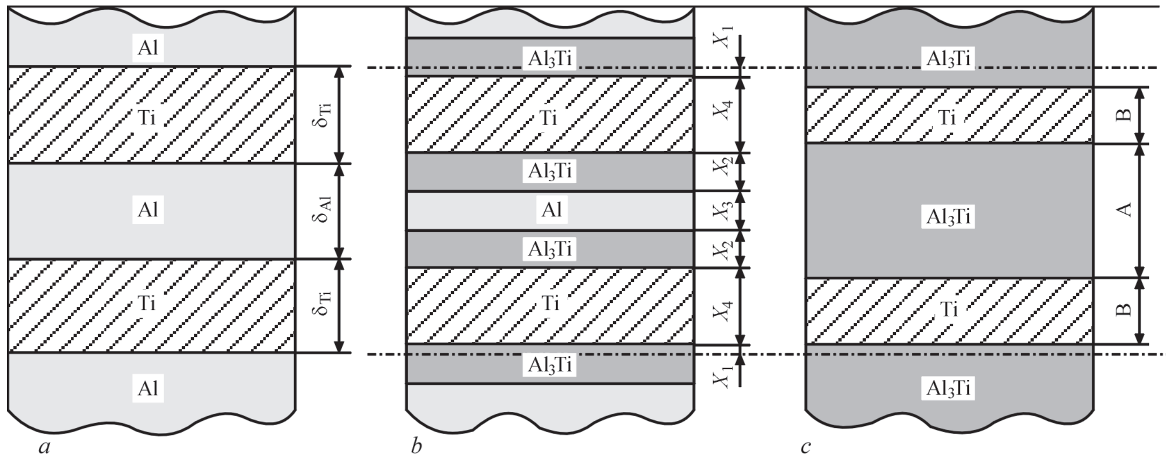


Fig re 3 Scheme of transformation of layers in a pack of Al and Ti foils: *a* — initial state; *b* — intermediate state; *c* – process completion [10]

of 630 °C and pressure of 25 MPa. Figure 3 gives the scheme of transformation of layers in a pack of Al and Ti foils. The work presents a procedure of calculation of the parameters of current processing of a multilayer pack to obtain metal-intermetallic Ti−Al_3Ti LCM. Selection of boundary values of current that allow reducing the time of pack processing without partial melting of the aluminium layers, was substantiated. For the considered case the current is 150 A.

One of the most widespread methods of LCM manufacturing is **rb ling**, which consists in joint rolling of a pack of metal layers.

The authors of [11] used the rolling method to obtain layered titanium-aluminium composite. For this purpose, they assembled a pack of plates from VT1-0 alloy (100×50×0.1 mm) and AD1 alloy (100×50×0.02 mm). The following thickness ratios of titanium and aluminium layers were selected: 100/20, 100/40, 100/60, 100/200 (μm). Rolling was conducted in steel sheaths at the temperature of 450–520 °C under a vacuum of 0.01 Pa. At the first passage the degree of deformation was equal to 25–40 %, at the following passes it was 8 to 10 %. As a result strips of 0.35 mm thickness with 4800 layers were produced. Here, the average thickness of the titanium layer was 100–120 nm, and that of the aluminium layer was 80–90 nm. Deformation of aluminium layers was by 25–30 % greater than that of titanium layers. It led to violation of the continuity of the titanium layer. Deformation of the subsurface layers of both the materials was by 10 to 15 % greater than that of the central layers. Experimental results allowed establishing an optimum rolling temperature (470 °C) and degree of cogging in rolling (up to 30 % in one pass), which allow welding the Ti and Al layers, and prevent formation of more than 0.5 % of intermetallic phases in the composite.

In work [12] a pack of two foils from AA1235 aluminium 300 μm thick and one titanium foil 25 μm thick was rolled at room temperature. The total thickness of the pack was reduced from 625 to 130 μm. Here, the titanium layer was broken up and dispersed in the aluminium matrix. This was followed by annealing the samples at 600 °C for 3600–10080 min with the purpose of formation of Al_3Ti intermetallic. At less than 720 min duration of heat treatment pores are observed (Kirkendale effect) on Al/Al_3Ti interface, because of a significant difference in Al and Ti diffusion. Pore presence leads to lowering of the values of ductility and strength of the composite sheet. Increase of annealing time to 1440 min promotes healing of pores, composite material consists of coarse-grained particles of Al and ultradispersed Ti and Al_3Ti. In this case also higher values of mechanical properties (σ_y = 135 MPa) are achieved. At annealing duration of 2880 min, pure Ti disappears with its complete transition into Al_3Ti. At further increase of annealing time to 10080 min, Al_3Ti particles are redistributed in Al matrix closer to LCM surface that leads to lowering of strength and ductility.

The authors of work [13] joined Al and Ti foils by the method of rolling at room temperature. After scraping by a metal brush and in acetone, the foils were assembled into a pack of 100×50 mm that consisted of layers of titanium 82 μm thick and aluminium 80–140 μm thick. It was found that the deformation level is almost the same for all LCM layers. However, bonding of the pack central layers occurs earlier than that of the external ones, which is caused by elastic recovery of the external layers that leads to breaking up of the bonds formed after passing under the rolls. Presence of scratches on aluminium after scraping leads to increase of the coefficient of friction on layer interfaces that promotes an increase of adhesion strength at the initial stage of the process and reduction of soft metal pressing out on the pack edges. Experimental results also showed that when a thinner aluminium layer (80 μm) is used, its thickness decreases significantly, compared to the initial value. Increase of Al thickness allows producing LCM with a more uniform change of relative thickness of both the metals, and aluminium extrusion is minimized (Figure 4).

Work [14] is a study of dislocation formation in aluminium at long-term annealing of Al/Ti/Al LCM. AA1235 aluminium foil 300 μm thick and pure titanium 25 μm thick were used for producing LCM. Cold rolling of a pack of three foils was performed in a four-roll mill with 50 mm diameter rolls. Total thickness of the pack was reduced from 625 to 130 μm. Rolling was followed by annealing of the produced joints at 600 °C for 360–10080 min. Heat treatment promotes Al and Ti diffusion to the interface and a reaction between them with formation of Al_3Ti intermetallic. Kirkendale effect and difference in molar volumes of the reagents results in pore formation on Al and Al_3Ti interface. Increase of annealing duration to 2880 min leads to a significant decrease of pore size. At heat treatment for 10080 min, anomalously high dislocation density of $7.5\cdot10^{14}$ m^{-2} is observed in Al layer near Al_3Ti intermetallic. The authors state that this is caused by development of pores and diffusion of Ti atoms. Interdiffusion of Ti and Al and Al_3Ti formation at long-term annealing of strongly deformed material allows some Ti atoms going beyond Al_3Ti stoichiometry limits and forming a buffer zone that consists of Ti solid solution in Al. With longer annealing time, large pores separate into smaller ones and are distributed in the aluminium matrix. Movement of atoms at a high temperature promotes gradual decrease of spherical pore size, which become vacancies as a result of it. In addition, in order to minimize free energy, the distributed Ti atoms were gathered in vacancies, and dislocations were formed and pinned by Ti atoms and clusters.

Ultrasonic welding belongs to solid-phase methods and envisages use of ultrasonic oscillations as the energy source, alongside pressure application. This process allows joining a wide range of metals up to 3 mm thick. Welding takes place at low temperatures that allows producing the joint without intermetallic formation.

The authors of [15] performed ultrasonic welding of an Al/Ti pack into LCM. A substrate from AA1100 aluminium alloy 1.5 mm thick was used as the first layer. Foil from pure titanium and AA1100 aluminium 127 μm thick was selected for the experiments. The foils were alternately assembled into a pack of 2–10 layers. Up to four foils were joined in one pass. Welding was followed by heat treatment at 480 °C for 240 min. It was found that rupture strength rises with increase of the total number of layers and reaches $\sigma_t =$ $= 216$ MPa for LCM of 10 layers that is attributable to

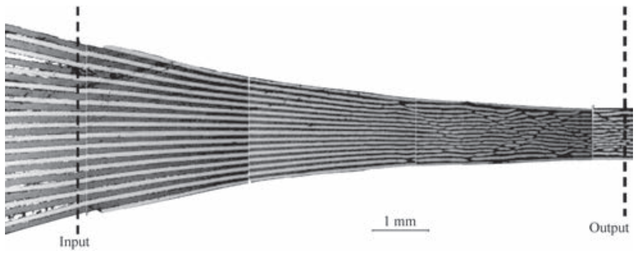


Figure 4 Cross-section of Al/Ti LCM produced by the rolling method [13]

increase of volume fraction of titanium. Layered material ductility also rises with increase of the number of layers up to 6 ($\delta = 32$ %), and after that it almost does not change.

Work [16] was a continuation of previous studies. Welding, however, was performed already without backing. Pure titanium and aluminium of 1100 grade up to 127 μm thick were used in the experiments. Foils were alternately assembled in packs with the total number of 7 to 15 layers, the first and last of which were from titanium. Samples were joined with a certain force and at sonotrode frequency of 20 kHz. Welding was performed in several stages, depending on the number of layers. First a pack of three foils of Al/TI/Al was assembled and passed under the sonotrode. This was followed by adding two more foils of Al + Ti and the process was repeated. Then annealing at 480 °C was performed for 240 min, in order to improve the joint quality. Increase of the number of layers has a negligible influence on mechanical characteristics of the joints. However, better results were obtained for a pack of 15 foils that is attributable to a greater number of passes in welding under the sonotrode.

The authors of work [17] studied the features of LCM joint formation under the impact of ultrasonic welding. Three foils from pure titanium and three from aluminium 1100 of 127 μm thickness each were used for this purpose. Welding was performed in Fabrisonic SonicLayer 4000 system of 9 kW power at the following parameters: 3500 N compressive force, 41.55 μm vibration amplitude, and 25.4 mm/s welding speed. In order to increase the ductility and intensify the adhesion process, the substrate from aluminium alloy 6061 was heated up to 93.3 °C. One titanium foil and one aluminium foil were welded in one pass. Discontinuities are observed on the interfaces between the individual layers. There are no intermetallics in the joint zones. During welding, mainly aluminium is deformed, that is indicated by grain disorientation by 9°, and more than two times reduction of their size at almost unchanged values for titanium. A conclusion was made that the main mechanism responsible for joint formation in ultrasonic welding is strong shear deformation on the layer interface.

Further studies were devoted to determination of the impact of heat treatment on the joint and they are covered in work [18]. The alloys and parameters of ultrasonic welding modes were the same as in the previous work. Joint samples were subjected to annealing at 600 °C for 60 min. Sound joints were produced, where considerable growth of grains in the aluminium layer and formation of a thin intermetallic interlayer on Al/Ti interfaces took place. Delamination testing results showed that the force required for sample breaking after heat treatment, is almost two times higher than that directly after welding (5.8 kN against 2.4 kN). Here LCM breaking after welding takes place within one layer, and in the annealed samples it runs through several layers that is indicative of higher joint strength along the layers, compared to the initial materials. Shear strength studies showed the same tendency of values — τ_{sh} = 46.3 MPa, after welding and τ_{sh} = 102.4 MPa for heat-treated LCM.

In work [19], the impact of heat treatment on mechanical properties of LCM produced by the method of ultrasonic welding was studied. Foil from 3003-N18 aluminium alloy 150 μm thick and pure titanium 75 μm thick was used. LCM was built up layer-by-layer on a massive substrate from 3003-N18 alloy, starting from titanium. The mode of welding aluminium foil was as follows: compressive force of 1750 N, vibration amplitude of 16 μm, welding speed of 23.7 mm/s; for titanium foil it was: compressive force of 2000 N, vibration amplitude of 28 μm; welding speed of 10.58 mm/s. Preheating temperature was 150 °C in both the cases. Then annealing at 480 °C was conducted for 30–270 min. The method of optical metallography showed continuous contact between Al and Ti, and no defects were found in the joint zone. Heat treatment for 30 min allows relieving the stresses, that arose between the metal layers during welding, and almost two times increasing the shear strength, compared to unannealed samples (τ_{sh} = = 72.96 MPa against 37.78 MPa). Increase of soaking duration at annealing leads to recrystallization and growth of the grains, resulting in a drop in strength.

Application of **diffusin welding** allows at the initial stage of the process localizing plastic deformation of matrix material, required for filling the space between the layers of the strengthening phase and for maximum densification of the matrix proper. When tight contact occurs between the layers, diffusion processes run on the interface, which provide strong bonding on these interfaces, and required strength level of the CM proper.

The authors of work [20] joined by diffusion welding the foils from TA1 titanium and 1060 aluminium alloys of 220×220×0.1 mm size. First 1 MPa pressure was applied to the samples in a vacuum furnace, to ensure tight contact between Al and Ti layers. Then, heating up to 500 °C was conducted for 180 min with soaking for 60 min at this temperature, to ensure uniform heating of the foil. Then temperature was increased up to 550 °C for another 30 min, and samples were soaked for 180 min to achieve diffusion welding of the layers, here pressure was increased up to 3 MPa. Finally, the foil was cooled together with the furnace under pressure of 1 MPa. Thus, LCM without pores or intermetallic inclusions on Al and Ti interfaces was produced (Figure 5).

Work [21] indicates the need for prior degassing of the foil in LCM production. The authors used foil from AM5 aluminium and VT1-0 titanium of thicknesses of 50 and 100 μm, respectively. Foil was rinsed in acetone, alternately assembled into a pack and heated up to 640 °C for 30 min in the vacuum of 1.333 Pa to ensure degassing. Rather intensive gas evolution is indicative of vacuum deterioration at this stage by an

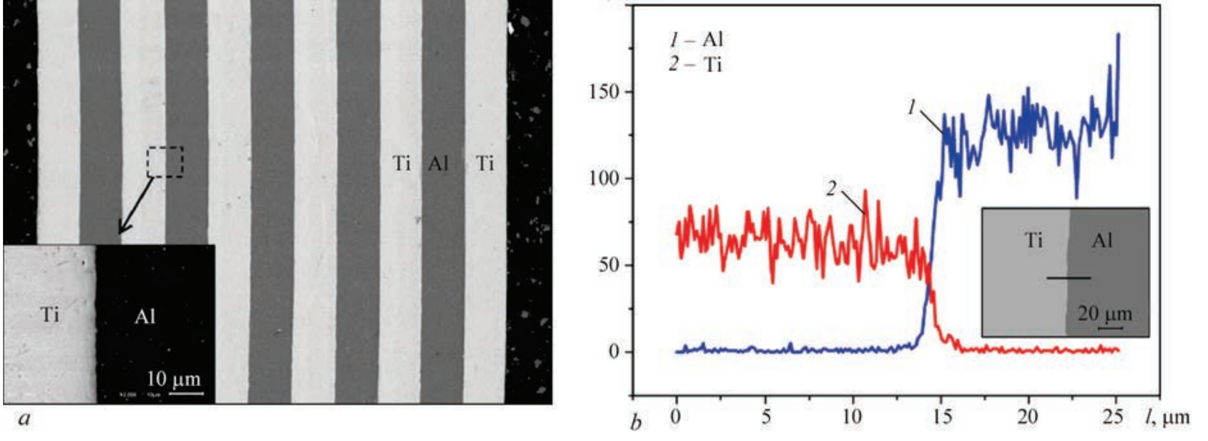


Fig re 5 Cross-section (*a*) and distribution of elements in the joint zone (*b*) of Al/Ti LCM, produced by the method of diffusion welding [20] order to 13.33 Pa. Then, the samples were compressed with a certain force and soaked at the same temperature for another 240 min. It resulted in producing Ti/Al_3Ti LCM. Here, a large number of microcracks and pores in place of the former aluminium foil are observed in the intermetallic layer.

In work [22], the method of diffusion welding was used to produce LCM from 1060 aluminum alloy 100 μm thick and TA1 titanium alloy 150 μm thick. Welding was performed in the following mode: 550 °C temperature, 5 MPa pressure, 180 min process duration and 10^{-3} Pa vacuum. To endure tight contact of the foils, pressure was applied before heating. Cooling to room temperature was performed together with the furnace. LCM without pores, discontinuities or intermetallic inclusions with ultimate strength σ_t = 224 MPa and relative elongation δ = 35 % was produced. Results of bend testing showed that microcracks on Al/Ti interface appear at bend angle of 120°. No sample bending occurs even at its bending up to 180°.

The authors of [23] studied the impact of thickness of aluminium and titanium foil on LCM formation by diffusion welding. Pure titanium (0.15–0.4 mm thick) and AA1060 aluminium (0.1–0.4 mm thick) were used as investigation materials. Al and Ti foil was cut up into 300×300 mm squares. Initial thickness ratio of Al and Ti layers was as follows: 0.1/0.15, 0.2/0.25, 0.4/0.4. Total pack thickness was 1.15–1.2 mm. Before welding, Ti was etched in a solution, consisting of 35 % nitric acid, 5 % fluoric acid and water. Aluminium foil was first cleaned in alkali solution of 30 g/l sodium hydroxide heated up to 50–60 °C, which was followed by immersion in an aqueous solution of nitric acid of 300–350 g/l at room temperature. Then samples were washed in water, then in ethanol. Diffusion welding was conducted in the following mode: 550 °C temperature, 5 MPa pressure, and 180 min soaking under the conditions of 10^{-2} Pa vacuum. Cooling to room temperature was performed in the furnace. Metallographic investigations showed that at reduction of the layer thickness the interface between them becomes wavelike. It results in increase of the total area of the surface of contact between the metals, compared to the initial condition. Reduction of layer thickness ratio from 0.4/0.4 to 0.1/0.15 also promotes increase of the diffusion zone from 4.1 up to 5.2 μm. Samples with thickness ratio of 0.4/0.4 (σ_t = 288.03 MPa) had the maximum tensile strength, however delamination between titanium and aluminium is absent only in 0.1/0.15 samples (σ_t = = 223.67 MPa).

In work [24] previous studies were continued with a focus on investigation of the process of initiation and propagation of a fatigue crack in LCM. Al and Ti foils of 0.15 and 0.1 mm thickness, respectively, were used. Diffusion welding was performed in the following mode: 500–600 °C temperature, 5 MPa pressure, 180 min soaking duration, and 10^{-3} Pa vacuum. Sound joints without intermetallic inclusions and with higher mechanical properties were produced at welding temperature of 550 °C (σ_t = 224 MPa). After 67554 loading cycles fatigue cracks appear on sample edges. Their initiation mechanism is considerable local plastic deformation resulting from high stress concentration. Fatigue cracks propagate mainly normal to the loading direction. During their propagation, large triangular cracks form, and fatigue cracks spread in several directions, forming a large number of microcracks. This is followed by interphase delaminations, which prevent further initiation and propagation of the fatigue cracks and release the fatigue cycle stresses. Threshold value of fatigue crack propagation in Al/Ti LCM was close to 7.4 MPa·m$^{1/2}$ that is higher than that in titanium and aluminium.

Work [25] is a study of the process of initiation and growth of Al_3Ti intermetallic phase between Al and Ti layers in diffusion welding with further annealing. Used for this purpose were aluminium and titanium sheets of 0.5 mm thickness. Before welding the samples were cleaned in an ultrasonic bath in 10 % solution of hydrofluoric acid. Aluminium and titanium sheets were stacked alternately up to the total thickness of 5 mm, pressed together with 4 MPa force, heated up to 550–575 °C and soaked for 120–360 min. Annealing was conducted at the welding temperature for 720–2880 min. After welding of 360 min duration a thin Al_3Ti interlayer forms in the joint zone, the thickness of which becomes greater during subsequent annealing. After heat treatment at 575 °C for 2880 min the aluminium layer disappears completely. It was found that at annealing at 575 °C for 2160 min Al diffusion rate is 20 times higher than that of Ti (D_{Al} = 27.133·10^{-15} m^2/s against D_{Ti} = = 1.202·10^{-15} m/s). It leads to pores formation near Al/Al_3Ti interface, as a result of Kirkendale effect. Such predominating diffusion of Al towards Ti stimulates initiation and growth of the intermetallic phase mainly on Ti/Al_3Ti interface. Figure 6 gives the scheme of Al_3Ti growth under the impact of temperature. During welding diffusion of individual Al and Ti atoms takes place with formation of Al_3Ti phase along Al/Ti interface (Figure 6, *b*). Annealing is accompanied by nonuniform growth of the intermetallic phase (Figure 6, *c*): ultrafine Al_3Ti (A) grains form near Ti, grains with coarse structure (B) form in the central part of Al_3Ti layer, and relatively small Al_3Ti grains (C) appear near the aluminium layer. This is caused by a high level of aluminium diffusion that promotes more in-

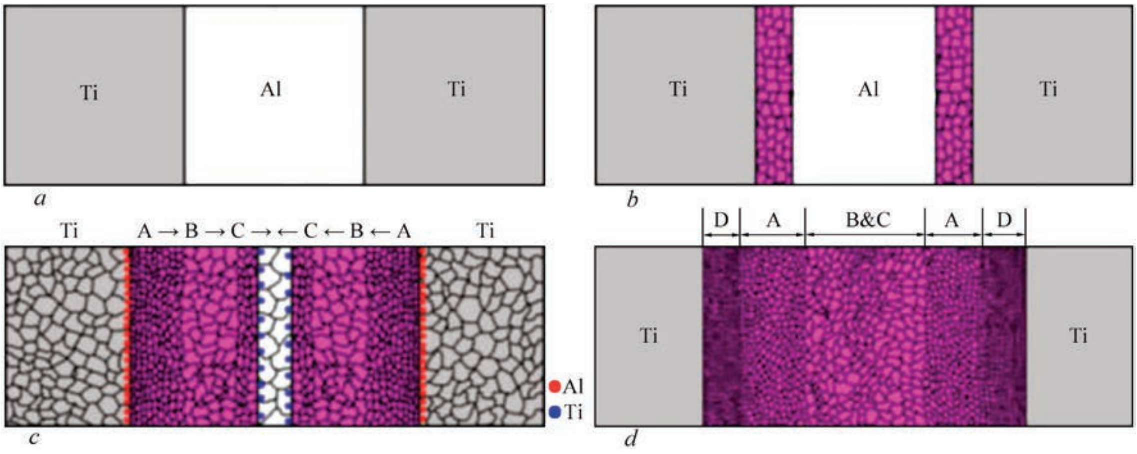


Figure 6 Schematic image of the process of Al_3Ti growth in Al/Ti layered system: *a* — initial stage; *b* — intermediate state; *c* — final structure after Al transition into intermetallic [25] (for A, B, C, B description see the text)

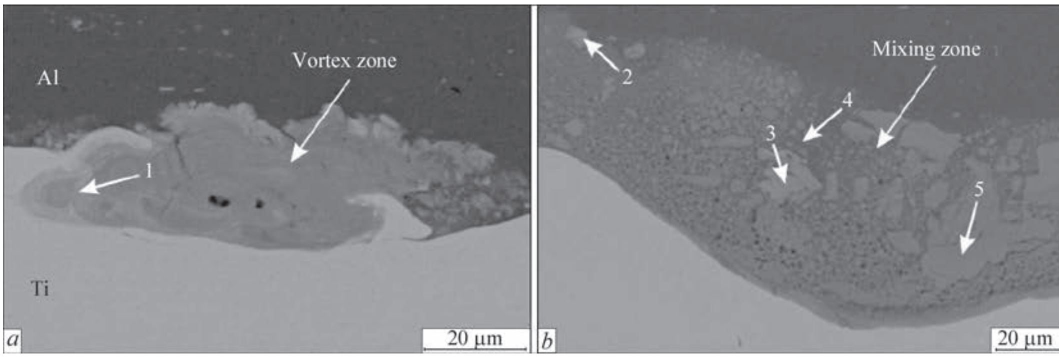
tensive formation of intermetallic phase nuclei on Ti/Al_3Ti interface with shifting of regions A, B, and C towards the aluminium layer. After complete transition of Al into the intermetallic, regions B and C mix in the central part of the intermetallic layer (Figure 6, *d*). Thus, the final structure of Al_3Ti grain consists of recently grown fine Al_3Ti grains on Ti/Al_3Ti interface (D), region A with a certain coarsening of the grains is shifted towards Al, that is adjacent to region D and mixed regions B and C in the central part.

The authors of work [26] studied reaction diffusion in a multilayer Al/Ti pack. Experiments were conducted with a pack of 17 titanium and aluminium foils of 90×25 mm size. Foils from Al and Ti were first etched in 10 % solution of hydrofluoric acid and 15 % solution of sodium hydroxide with further cleaning with alcohol and rinsing in water. Joining was performed at 600 °C temperature, 50 MPa pressure for 180 min under a vacuum not lower than 10^{-2} Pa. Then the produced samples were annealed at 520–650 °C for 60–9000 min. It was found that at heat-treatment in the selected temperature range Al_3Ti is a single phase that is observed in the diffusion zone. Titanium and aluminium diffuse in the direction towards each other. Al_3Ti layers, however, grow mainly towards Al. At 575 to 600 °C temperature, the kinetics of Al_3Ti growth changes from the parabolic to linear one. In low-temperature kinetics, diffusion of Ti atoms along the grain boundaries of Al_3Ti layers prevails, whereas the reaction on the Al/Al_3Ti interface in the high-temperature mode is limited by diffusion of Ti atoms into Al foil on Al_3Ti growth front, as a result of higher solubility of Ti in Al with temperature rise.

In work [27] the impact of titanium thickness on formation of Al/TI LCM was studied. Titanium of 45–180 μm thickness and aluminium of 54 μm thickness were used. Foil of 40×80 mm size was scraped with a wire brush and cleaned in alcohol in an ultrasonic bath. Then 50 Ti layers and 49 Al layers were alternately stacked in a graphite matrix. Welding was conducted in a vacuum furnace in the following mode: 500 °C temperature, 50 MPa pressure, soaking for 30 min. Then, the produced sample was annealed without applying pressure at 900 °C for 30 min for complete transition of aluminium into the intermetallic phases. Final sintering was performed in high vacuum of the order of 10^{-3} Pa at 1050 °C, under a pressure of 50 MPa for 60 min, heating rate was 10 °C/min, and cooling rate was about 15 °C/min. After all the stages of thermodeformational impact, joints with slight porosity were produced. Irrespective of titanium thickness, the following sequence of phase layers formed: α-Ti, $AlTi_3$, AlTi, Al_2Ti and Al_3Ti. Samples produced with titanium of 90 μm thickness had the highest tensile strength (σ_t = 606 MPa) and fracture toughness (K_{1C} = 47.6 MPa·m$^{1/2}$). This is attributable to optimum ratio (almost one to one) between α-Ti and total thickness of intermetallic layers. Such a volume fraction of α-Ti layers effectively prevents continuous crack displacement in the opening mode, making it deviate and bifurcate in the intermetallic layers. Plastic deformation of titanium may absorb a large amount of energy up to material fracture.

Explosion welding has limited application for LCM production. It is solid-phase process of producing the joints with intensive plastic deformation of metal under the impact of a short high-amplitude pulse.

In work [28] explosion welding was applied to produce Al/Ti LCM. VT1 titanium and aluminium (Al–1Mn) alloys of 100×200×0.2 mm and 100×200×0.25 mm size, respectively, were used as research materials. The sheets were alternately placed into a pack with the total number of 40 pcs. Ammonite powder 6GV of 0.9 g/cm^3 density and 4200 m/s detonation rate was used as explosive. Further heat



Spectrum	Weight, %	
	Al	Ti
1	59.5	40.5
2	61.0	39.0
3	65.5	34.5
4	86.0	14.0
5	74.0	26.0

Figure 7 Structure and chemical composition of mixing zones formed after explosion welding of Al/Ti LCM: *a* — vortex; *b* — dispersed inclusions (marked by arrows) [28]

treatment was conducted in air atmosphere at the temperature of 640 °C, and pressure of 3 MPa for 1200 min. In the produced samples a change of the joint zone between the aluminium and titanium layers from the wavelike to almost straight one is observed with greater distance from the explosion epicenter. Nonuniformity of the impact of explosion wave is the cause for presence of micropores between the individual layers. The following intermetallic phases form in the mixing zones: Al_3Ti, AlTi, Al_5Ti_3, and $AlTi_3$ (Figure 7). Further LCM heat treatment led to growth of Al_3Ti, the growth rate of which was four times higher, compared to samples made by the sintering technology. Pressure application at annealing allows producing samples with minimum number of defects.

Authors of works [29, 30] used explosion welding to produce a pack of six aluminium and titanium sheets 0.5 mm thick. A mixture of ammonium nitrate with trotyl and gas oil 20 mm thick was used as the explosive. The detonation rate was approximately 4500 m/c. Then LCM samples were annealed at 630 °C for 60–15600 min in ambient atmosphere. After welding the aluminium and titanium joint zone is a straight line without vortices. In some places, regions with Al_3Ti are observed between Al and Ti, formation of which is related to local temperature rise in welding. During the first 780 min at annealing the rate of intermetallic layer growth is of a linear nature, and after that its behaviour changes to a parabolic one. This is indicative of transition of Al_3Ti growth mechanism from controlled reaction to controlled diffusion.

In work [31], Al/Ti bimetal was produced from rather thick sheets by explosion welding method. Plates from Ti Gr. 2 titanium (140×460×0.8 mm) and A1050 A aluminium (140×460×4 mm) were used for this purpose. The distance between the metals before the explosion was 1.5 mm, and the detonation rate was 1900–1950 m/s. Then, the samples were annealed in sealed quartz ampoules at 552 °C for 30–6000 min. After welding sound joints were produced. The zone of aluminium and titanium joint acquires a wavelike appearance, in the vortices of which Al_3Ti, Al_2Ti, AlTi and $AlTi_3$ intermetallics form. The annealing process makes Al_3Ti phase the dominating one, promoting growth of its layer over the entire joint surface. Investigations of the kinetics of Al_3Ti intermetallic revealed four stages of its growth: incubation period (up to 90 min); growth controlled by a chemical reaction (90–300 min), mixed mechanism of the chemical reaction and volume diffusion (300–2160 min); and volume diffusion (2160–6000 min). Intensive growth of aluminium grains occurs during heat treatment.

Conclusions

Based on the presented review of publications, we can state that:

1. Application of LCM can greatly improve the properties of structures made from them, namely fracture toughness, fatigue and impact characteristics, wear, corrosion and damping ability.

2. Process of joining aluminium to titanium can be conducted both under vacuum, and in air. Here, depending on the final product, it is possible to produce LCM both with strengthening due to formation of intermetallic interlayers, and without it.

3. The main methods of producing LCM, based on aluminium and titanium is melt impregnation, rolling, sintering, ultrasonic, explosion and diffusion welding.

4. Each welding process has its characteristic features. The melt impregnation method envisages contact of liquid aluminium with solid titanium and does not allow producing LCM without the intermetallic interlayer. The rolling process requires considerable plastic deformation of the pack to obtain the laminates that leads to a considerable number of the rolling cycles. Ultrasonic welding is used only to produce a layered pack of thin foils. Explosion welding is accompanied by formation of a wavelike joint zone and requires further heat treatment. Diffusion welding, depending on welding mode parameters, allows producing LCM both with formation of the intermetallic phase in the butt joint, and without it.

Proceeding from the possibilities of wide adjustment of LCM structure and composition at the stage of forming the joint by diffusion welding, it is urgent to perform further investigations on production of such materials with controlled content of the intermetallic phase in the butt joint.

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Received 29.01.2020