



# Article Ta/Ti/Ni/Ceramic Multilayered Composites by Combustion Synthesis: Microstructure and Mechanical Properties

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**Abstract:** Ta/Ti/Ni/ceramic multilayered composites were successfully prepared by combustion synthesis. Laminated composites Ti–Ta–(Ti + 0.65C)–Ni–(Ti + 1.7B)–(Ti + 1.7B)–Ta–Ni-Ti and 3(Ti + 1.7B)–Ta–(5Ti + 3Si)–Ta–(Ti + 1.7B)–Ta–(5Ti + 3Si)–Ta–(Ti + 1.7B)–Ta–(5Ti + 3Si)–Ta–(Ti + 1.7B), and (2) reactive tapes (Ti + 0.65C), (Ti + 1.7B), and (5Ti + 3Si), which, upon combustion, yielded ceramic layers as starting materials. The microstructure, crystal structure, and chemical composition of multilayered composites were characterized by SEM, EDX, and XRD. Their flexural strength was measured at 1100 °C. Upon combustion, Ta foils turned strongly joined with Ti ones due to the development of high temperature in the reactive layers yielding TiC<sub>x</sub> and TiB<sub>y</sub>. The formation of a liquid phase between metallic foils and reactive tapes and mutual interdiffusion between melted components during combustion favored strong joining between refractory metallic foils. Good joining between metals and ceramics is reached due to the formation of thin interfacial layers in the form of cermets and eutectic solutions.

Keywords: multilayered metal-ceramic composites; combustion synthesis; reactive tapes; Ta; Ti

# 1. Introduction

Layered composite materials (cermets) that combine the properties of metals (ductility, heat resistance, and thermal conductivity) and ceramics (hardness, corrosion resistance, high-temperature strength, and low thermal conductivity) find wide application in different areas of the aerospace industry, mechanical engineering, power engineering, etc. Meanwhile, the fabrication of multilayer composite materials encounters serious difficulties caused by strongly different mechanochemical parameters of metals and ceramics to be joined. In view of their brittleness and chemical inertness, engineering ceramics are hard to join with metals [1]. The good joining of these dissimilar materials often requires the use of high temperature and pressure [1,2], pre-processing of surfaces to be joined, and the use of brazing alloys [3–8].

To date, multilayer cermet materials can be fabricated by several methods such as brazing [7–10], solid-state joining via friction- and laser-assisted welding [11–13], physical vapor deposition [14], spark plasma sintering [15], ceramic–metal diffusion bonding [4,5], melt infiltration [16,17]), and field-activated, pressure-assisted synthesis [18].

Brazing is a process of joining upon heating up to some certain temperature or putting a filler metal (solder) into the joint, the filler metal having a lower melting point than the adjoining metal. The applicability of this technique is strongly restricted by the poor wettability of ceramics with most metals and alloys [7–10]. Solid-state joining routes afford for joining metals with ceramics without the formation of the liquid phase. In this case, the main drawbacks are expensive facilities, the use of high pressure, a large duration of applied pressure, and the restricted configuration of resultant items [11–15]. The methods



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**Copyright:** © 2021 by the authors. Licensee MDPI, Basel, Switzerland. This article is an open access article distributed under the terms and conditions of the Creative Commons Attribution (CC BY) license (https:// creativecommons.org/licenses/by/ 4.0/). of liquid-phase joining (ceramic-metal diffusion bonding, infiltration bonding) require the presence of the liquid phase, some certain optimal exposure time, and close contact between the adjoining surfaces [4,5,19]. Often lightweight metals (such as Al) are used as a transition layer between the metal and ceramic [17,20]. However, this can be accompanied by chemical reactions in reactive layers yielding in situ thick intermetallic layers that diminish the strength of resultant cermet composites [21].

Joining metals with ceramics is a complicated issue due to the strongly difficult physical properties of these dissimilar materials. The quality of joining is governed by the following two factors: (1) the fit of thermal expansion and (2) the formation of undesired interfacial layers [19]. The problem of joining dissimilar materials still remains to be of current importance despite the considerable number of publications in the field.

Combustion synthesis (also known as self-propagating high-temperature synthesis (SHS)) is a facile technique for the synthesis of materials and deposition of coatings, which is desirable due to a wide range of suitable reagents, relatively simple facilities, and high combustion temperatures developed during combustion reaction [22,23]. SHS method was found applicable to joining such dissimilar materials as ceramics, metals, and carbon materials [24–26].

In this study, we report on the SHS-assisted joining between Ti and Ta foils and ceramic layers SHS-produced in situ from reactive Ti–B, Ti–C, and Ti–Si tapes. According to [27,28], the reactive cold-rolled Ti–B tapes 100–300 µm thick are capable of rapid burning at a high rate of heat release.

Due to the unique properties of Ta–Ti alloys (corrosion resistance, ductility, mechanical strength, heat resistance, etc.), they are widely used in modern mechanical engineering, electronics, aerospace industry, and nuclear power engineering [29–31]. However, practical implementation of these alloys is often restricted by a high melting point of Ta, its large density, and the high production cost of Ta–Ti alloys. This work aimed at searching for a combustion-aided route for the preparation of new more lightweight and cost-effective multilayer Ta-based composites that would correspond to modern requirements to structural and functional cermet materials.

The basic novelty of this study is in the preliminary structurization of starting sandwichlike samples by compiling the stacks of metal foils and reactive tapes. Combustion in reactive tapes is expected to yield joining ceramic layers in between the metal foils. Reactive tapes and metal foils had a close thickness (230–290  $\mu$ m). Previously [32,33], we have found that an important prerequisite for successful combustion-aided joining between Ti and Ta foils with ceramic is (1) the presence of the liquid phase and (2) some congruency between the melting points of ceramic composition and metal. Accordingly, the choice of individual layers for compiling green sandwich-like multilayer assemblies was performed with due regard to key thermophysical parameters of foils and rolled reactive tapes presented in Table 1 [19].

Composition	$T_{mp}$ , °C	$T_{ad}$ , °C	lpha, 10 <sup>-6</sup> deg <sup>-1</sup>
Ti	1670	-	8.6
Та	3017	-	6.5
Ni	1453	-	13
Ti + 1.7BT	-	3180	7.4
Ti + 0.65C	-	2380	9.6
5Ti + 3Si	-	2130	12.5

**Table 1.** Melting points  $T_{mp}$  for foil metals, adiabatic combustion temperature  $T_{ad}$  for reactive tapes, and coefficients of thermal expansion  $\alpha$  for all constituent materials.

In this work, Ni was selected because (1) it shows a relatively low melting point, which must facilitate the formation of the liquid phase, and (2) it is capable of forming  $TiB_2$ –Ni cermets [34]. Moreover, the formation of a Ni binder doped with the grains of

titanium borides and titanium carbides can impart additional strengthening to a resultant composite [35].

#### 2. Materials and Methods

Commercial powders of Ti (PTS-1 brand, mean particle size 40  $\mu$ m, 99% pure, Polema JSC, Tula, Russia), carbon black (P804T, JSC YATU, named after V.U. Orlov, Yaroslavl, Russia), amorphous boron (99.4%, AVIABOR, Dzerzhinsk, Russia), Si (d < 10  $\mu$ m, 99.4%, IC Bourevestnik, Sankt-Peterburg, Russia) and foils of Ti (VT-0, 180  $\mu$ m thick, LLC, RedMetSplav, Ekaterinburg, Russia), Ta (TVCh, 100  $\mu$ m thick, LLC, RedMetSplav, Ekaterinburg, Russia), and Ni (NP2M, 100  $\mu$ m thick, LLC, RedMetSplav, Ekaterinburg, Russia) were used as starting materials.

Reactive Ti–C and Ti–Si and Ti–B tapes (290 or 230  $\mu$ m thick, see Table 2) were prepared by cold rolling on rollers from Ti + 0.65C, 5Ti + 3Si, and Ti + 1.7B powder mixtures, respectively [27,28]. In order to remove moisture and impurity gases, the rolled tapes were held in a vacuum furnace (10<sup>-2</sup> Pa) at 700 °C for 2 h.

**Table 2.** Layer-by-layer structure of starting sandwich Sets **1–3**; for the sake of brevity, layers 1–13 are presented here (and hereinafter) as L1–L13.

	Set 1	Set 2	Set 3
L1	Ti (180 μm)	Ti (180 μm)	Ti + 1.7B (230 μm)
L2	Ti (180 μm)	Ti (180 μm)	Ti + 1.7B (230 μm)
PL3	Ta (100 μm)	Ta (100 μm)	Ti + 1.7B (230 μm)
L4	Ti + 0.65C (290 μm)	5Ti + 3Si (290 μm)	Ta (100 μm)
L5	Ni (100 μm)	Ni (100 μm)	5Ti + 3Si (290 μm)
L6	Ti + 1.7B (230 μm)	Ti + 1.7B (230 μm)	Ta (100 μm)
L7	Ti + 1.7B (230 μm)	Ti + 1.7B (230 μm)	Ti + 1.7B (230 μm)
L8	Ta (100 μm)	Ta (100 μm)	Ta (100 μm)
L9	Ni (100 μm)	Ni (100 μm)	5Ti + 3Si (290 μm)
L10	Ti (180 μm)	Ti (180 μm)	Ta (100 μm)
L11	_	-	Ti + 1.7B (230 μm)
L12	_	_	Ti + 1.7B (230 μm)
L13	-	-	Ti + 1.7B (230 μm)

Green  $30 \times 12.5$  mm sandwich-like samples of Sets **1–3** were stacked using alternating layers of metallic foils and reactive tapes, as shown in Table 2. Depending on the number and thickness of selected layers, the stack thickness could be varied between 1.69 and 2.59 mm.

Combustion was carried out in a constant-volume reactor under 1 atm of Ar and a pressure of up to 3 MPa was applied to heating plates 3 (Figure 1). In experiments, samples 2 (Figure 1) were heated at a rate of 60 deg/min to  $530 \pm 30$  °C (below self-ignition temperature), heaters 3 were switched off, after which samples 2 were ignited with electrically heated coil 4.

Readings of thermocouples T1, T2, T3 were displayed via an analog-to-digit converter (ADC; L-780, L-Card, Moscow, Russia) on the screen of PC (Figure 1). For better contact, thermocouple heads were cold-rolled down to a thickness of 40  $\mu$ m. Thermocouples T1 and T2 were placed in between reactive Ti + 1.7B tapes (Figure 1 and Table 2) at a separation of 5 and 14 mm, respectively, from igniting coil 4 (covering zone L6–L7 in Set 1). Thermocouple T3 monitored a temperature between layer *10* (Set 1) and graphite plate 5 at a separation of 25 mm from coil 4 (Figure 1, Set 1).

In the case of Set **3**, thermocouples T1, T2, and T3 were all installed between L8 and L9 (Table 2) at a separation of 5, 14, and 25 mm, respectively, from coil 4.



**Figure 1.** Schematic of the experimental setup: (1) reactor, (2) sandwich-like sample, (3) heaters, (4) igniting coil, and (5) graphite plates; T1, T2, T3 thermocouples (WR5/WR20).

Burning velocities *U* were derived from the analysis of still frames of video records of 5000 frames per second (camera MIRO M310, Vision Research, Wayne, NJ, USA) and time spans between thermocouple readings with an accuracy of  $\pm 6\%$ . Notably, the absolute temperatures shown by thermocouples T1–T3 normally yielded lower estimates.

Combustion-synthesized composites were characterized by SEM (Vega 3 TESCAN, Brno, Czechia), and EDX (Oxford Inca spectrometer, Oxford, UK) using Aztec software and XRD (DRON-3M diffractometer,  $Cu-K_{\alpha}$  radiation, Burevestnik, St. Petersburg, Russia). The thermomechanical behavior of synthesized samples was determined by three-point loading at 1100 °C under Ar using a modified testing machine Instron-1195 (Instron Ltd., Buckinghamshire, UK).

#### 3. Results and Discussion

# 3.1. General Features of Combustion

Comparative analysis of still frames and thermocouples readings allowed us to conclude that, in Set 1, the combustion wave propagated in a mode of frontal combustion at a mean burning velocity (*U*) of around 195 mm/s. Thermocouples T1 and T2 (Figure 1) showed a temperature of 2900 °C (Figure 2) close to the adiabatic one for the system under consideration. All thermograms showed the presence of an isothermic plateau (0.01–0.05 s long) that is indicative of melting the reagents and products. The processes of product patterning were seen to occur within the post-combustion zone in the temperature range 2200–1800 °C. It follows that a rather short (about 0.2 s long) process of combustion synthesis could actually yield lightweight tough platelets, with dimensions of 32 mm × 14 mm × 1.5 mm.



Figure 2. Readings of thermocouples T1–T3 during combustion of Set 1 (Figure 1 and Table 2).

# 3.2. Sandwiches of Set 1

The diffraction pattern of combustion products formed on the side surface of Set **1** shows (Figure 3) the presence of *bcc* Ta (*Im3m*) [36], orthorhombic TiB (*Pnma*) [37], hexagonal TiB<sub>2</sub> (*P6/mmm*) [38], and hexagonal Ti–Ta eutectics (*P63/mmc*) [39].



Figure 3. Diffraction pattern for cross section of burned Set 1.

Our SEM/EDX results (Figure 4) confirmed the combustion-synthesized material represents a frame of Ta foils well joints by the interfacial Ti–Ta, Ti–Ni, and Ti–Ni–Ta layers. Despite a high temperature registered by thermocouples T1 and T2 (around 2900 °C), we observed no traces of Ta melting (m.p. 3017 °C). This can be associated with a high rate of wave propagation and heat sink into the environment. Elemental mapping in Figure 4 showed the distribution of Ta, Ti, and Ni over the cross section of Set **1**.



**Figure 4.** SEM image (**a**) and EDX-mapping (**b**–**d**) for the cross-section of the layered composite (layers indicated to the left, see Table 2) derived from Set **1**.

#### 3.2.1. Interlayer Boundaries and Ta-Ti Joint

In order to study the joining between metal foils (Ta, Ti, Ni) and ceramic layers (TiB, TiC), the microstructure and chemical composition of their interfaces were analyzed by SEM/EDX. (Figure 5).

Figure 6 (also corresponds to the Figure 5b) shows the microstructure of interfacial Ti–Ta joint formed between layers L1–L2 and L3 in Set 1. A smooth and clear-cut pore- and crack-free joint between Ta and Ti foils (Figure 6a) arises due to the formation of a thin (~1 µm) interfacial layer of Ti–Ta solid solution. Some waviness of the joint line can be associated with the thermally induced diffusion of Ti into Ta in tightly compressed layers, which is typical of interphase instability [40,41]. Due to its basket-type martensite structure, *bcc*  $\beta$ -Ti (Figure 6b) can be expected to improve the toughness of the composite [41]. In our case, the  $\alpha$ -Ti  $\rightarrow \beta$ -Ti phase transition in L1–L2 layers could occur in the course of combustion since their temperature exceeded 880 °C.



**Figure 5.** Microstructure of interlayer boundaries formed in cross section of burned Set 1: (**a**) a general view, (**b**) Ti–Ta interface at (L1–L2)–L3, (**c**) Ta–cermet interface at L3–(L4–L6), (**d**) (Ni + TiB<sub>x</sub>)–Ta interface at (L4–L7)–L8, and (**e**) Ta–Ni–Ti interface at L8–(L9–L10).



**Figure 6.** Microstructure of Ti–Ta joint at (L1–L2)–L3, cross section of burned Set **1** (Table 2): (**a**) SEM image of Ti-Ta layer; (**b**,**c**) EDX mapping of Ti and Ta, respectively.

#### 3.2.2. Ta-Cermet Interlayer

In contrast to previous Section 3.2.1, zone L3–(L4–L6) (Set 1, Figure 7, also corresponds to the Figure 5c) exhibits no distinct separation line between the Ta layer (L3), and the products formed upon combustion of reactive tapes L4 and L6 together with Ni layer (L5). The cermet layer formed upon combustion was found to contain the grains of TiB (cubic shape), TiB<sub>2</sub> (hexagonal shape), and TiC uniformly distributed over a Ti–Ni matrix [42]. As could be expected, the Ni foil completely melted, while the shape of TiB, TiB<sub>2</sub>, and TiC grains and their size (0.5–1.0  $\mu$ m) are indicative of reaction completeness [35]. The formation of such a layer must improve the strength characteristics of composite material [43].



**Figure 7.** Microstructure of Ta–cermet composite formed upon combustion in zone L3–(L4–L6), cross section of burned Set **1**.

A wavy structure of the Ta–ceramics joint (Figures 5c and 7), the absence of pores and cracks, and the formation of the Ta + Ni<sub>x</sub>Ta layer suggest that the joining proceeded by the mechanism of high-temperature diffusion. Complete consumption of Ni foil L5 favored the mutual interdiffusion of Ta and Ti, which is known to occur at 1100–1800 °C [44,45]. The formation of solid solutions Ni + Ni<sub>3</sub>Ta, Ni<sub>3</sub>Ta + Ni<sub>2</sub>Ta, and Ta + Ni<sub>2</sub>Ta that may improve the high-temperature strength of composites is possible within the range 1350–1790 °C [45,46]. Our thermograms in Figure 2 suggest that the above phases can only be present in amounts of below 5%, and accordingly, the diffraction pattern in Figure 3 does not show the presence of the above phases.

## 3.2.3. Ta-Cermet Joint

The microstructure of the Ta–cermet joint presented in Figure 8 (also corresponds to the Figure 5d, zone L5–L8, Set 1) illustrates the formation of well-pronounced transition layers without visible pores and cracks. In this case, the joint is seen to consist of three interlayers: between the cermet layer, consisting of TiB (cubic shape) and TiB<sub>2</sub> (hexagonal shape) grains, uniformly distributed in the Ti–Ni matrix, and the Ta + Ni<sub>x</sub>Ta solid solution layer, a Ti–Ni layer is formed.

The absence of a distinct separation line between the layers supports the occurrence of diffusion, while the formation of new phases provides evidence of chemical reactions. It is also worth mentioning that the thickness of intermediate layers (5–15  $\mu$ m) is markedly smaller than that of starting Ta and Ni foils (100  $\mu$ m).



Figure 8. Microstructure of cermet-Ta joint (zone L5-L8, cross section of Set 1).

#### 3.2.4. Ta-Ni-Ti Joint

Figure 9 (also corresponds to the Figure 5e) shows the microstructure of the Ta–Ni–Ti joint (zone L8–L10, Set 1). At the interface of Ta and Ni and Ti foils, we observed the melting of Ti and Ni foils and the formation of the Ti–Ni solution (Figure 9). This is accompanied by partial melting of the Ta foil and the diffusion of Ta atoms into the newly formed Ti–Ni solution.



Figure 9. Microstructure of Ta-Ni-Ti interface (zone L8-L10, cross section of burned Set 1).

In this case, strong joining is reached due to the occurrence of reactive diffusion, mutual interpenetration, and chemical reactions in reactive tapes. The situation is close to the mechanism of ceramic–metal diffusion bonding: strong joining requires the presence of the liquid phase existing a sufficiently long time [5,6]. However, in contrast to ceramic–metal diffusion bonding, our process requires no complicated equipment, appears advantageous in processing time, and allows for avoiding the formation of the thick transition layers that may decrease joint toughness.

According to [47,48], strong bonding of dissimilar materials by ceramic–metal diffusion bonding can be reached when the process temperature attains 0.5–0.7 of metal melting point. Another prerequisite for successful combustion-assisted joining of metal with ceramic is compliance between a reaction temperature and melting point of foil materials. In other words, the joining of metals with ceramics requires rather high temperatures that can be reached upon the proper choice of reactive tape composition.

Important advantages of our approach are (1) the use of relatively low load P (Figure 1) and (2) no need for the elimination of oxide films from the surfaces to be joined and their polishing since the oxide films decompose at high combustion temperatures.

#### 3.3. Set 2

The only difference between Set **1** and Set **2** is the use of 5T + 3Si mix instead of Ti + 0.65C one in position L7 (Table 2). Due to a lower temperature ( $T_{ad} = 2130 \text{ °C}$ , Table 1), we failed to achieve strong joining in case Set **2**. Nevertheless, we used this system in control experiments on the impact of temperature on the combustion synthesis in the systems under consideration.

An insignificant decrease in reaction temperature (by 200-250 °C) caused by the use of 5T + 3Si reactive tapes in Set **2** made reliable joining of foils with combustion-synthesized ceramics impossible (Figure 10).





## 3.4. Set 3

The structure of the sandwiches of Set **3** based on the use of Ti–Si reactive tapes is presented in Table 2. Figure 11 shows the readings of thermocouples T1–T3 (all installed between layers L8 and L9) during the combustion of Set **3** (Figure 1, Table 2). The *T* values in Figure 11 are close to those reported for 5Ti + 3Si mixture previously [32,33]. These results were used to determine the burning velocity in Ti–Si reactive tapes: U = 45-50 mm/s. Compared with Set **1**, this is smaller by a factor of around three. Nevertheless, this fact did not prevent us from obtaining a good joint between Ta foils and titanium diboride.



Figure 11. Readings of thermocouples T1–T3 during combustion of Set 3 (cf. Figure 1 and Table 2).

Figure 12 presents the fragment of cross section for a composite derived from Set 3, and Figure 13 shows its diffraction pattern. In this case, we managed to obtain a good joining of Ta foils with the newly formed TiB<sub>2</sub> ceramics (Figure 12). The diffraction pattern in Figure 13 shows that the synthesized material contains *bcc* Ta (*Im*3*m*) [36] and TiB<sub>2</sub> (*P*6/*mmm*) [38].



Figure 12. Cross section fragment of a composite derived from Set 3 (Table 2).



Figure 13. Diffraction pattern for the cross section of composite derived from Set 3.

Figure 14 illustrates the microstructure of the Ta–TiB<sub>2</sub> joint in Set **3**. It is seen to consist of thin (3–5  $\mu$ m) interlayers of (Ti,Ta) solution and interstitial (Ti,Ta)B phase. Partial substitution of Ti by Ta atoms becomes possible due to diffusional penetration of Ta atoms into the ceramic layer. According to [49], Ta atoms are capable of penetrating, at 1000 °C, to a depth of 11  $\mu$ m in 10 min. In this case, the newly formed (Ti,Ta)B phase exhibited a microhardness of up to 30 GPa [49]. Therefore, despite some porosity of the transition layer (Figure 10), it is the formation of the (Ti,Ta) and (Ti,Ta)B phases that ensure the sample cohesion and integrity.



Figure 14. Microstructure of Ta–TiB<sub>2</sub> joint at L7–L8 interface in cross section of burned Set 3.

As can be judged from Figure 15, strong joining with Ta was reached due to the mutual interdiffusion of Ta and Si atoms. The temperatures shown by T2 and T3 (Figure 11) are



above the melting point of Ti (1670  $^{\circ}$ C) and Si (1414  $^{\circ}$ C), which ensures the presence of the liquid phase.

**Figure 15.** Microstructure of  $(Ti,Ta)Si_x$  joint formed between Ta layers L8 and L10 (cross section of burned Set **3**).

In conditions of diffusion bonding (1000–1500 °C), Ta and Si atoms were found to cover a distance of up to 30 µm from a Ta substrate [49,50]. In this case, the combustion-aided joining of Ta foil with Ti–Si ceramic was reached due to the formation of (Ti, Ta)Si<sub>x</sub> solid solutions, although the Ti<sub>x</sub>Si<sub>y</sub> and Ta<sub>x</sub>Si<sub>y</sub> could not be detected by XRD. Our Figure 15 demonstrates the formation of the grained structure formed by (Ti,Ta)Si<sub>x</sub> compounds uniformly joined with Ta layers L8 and L10 without any visible interfacial phases. Such a joint promises good strength properties of synthesized materials at high temperatures.

#### 3.5. Thermomechanical Behavior

Three-point bending tests were carried out on an upgraded universal testing machine Instron-1195 at a temperature of 1100 °C in an argon atmosphere. We monitored the graphical load-deflection in real time on the PC screen and recorded the observed vertical deflection  $\Delta$  in the form of the *P*( $\Delta$ ) function that could be recalculated into the  $\sigma(\Delta)$  function. The load rate was 0.5 mm/min.

Samples derived from Set 1 showed good flexural strength (up to 128 MPa at 1100 °C) and Young modulus (around 35 GPa). The flexural strength of samples derived from Set 3 was markedly lower (72–73 MPa). This can be associated with the formation of thick  $TiB_2$  layers at the lateral sample edges. Mechanical properties of our composites derived from Set 1 and Set 3 turned higher than those (74 MPa at 800 °C) of layered  $TiB_2$ –TiC–SiC/Ta cermets reported in [49]. Materials obtained in this work are characterized by the same strength values (Set 3) and higher (Set 1), but in contrast to [49], these values were obtained at 1100 °C.

In order to improve the strength characteristics of cermets derived from Ti–Si-containing reactive tapes, the upper and lower layers of a sandwich should be foils. Moreover, we have to prevent the formation of thick ceramic layers. This can be reached through (1) proper interleaving reactive tapes and metal foils, (2) the use of additional low-melting foils (such as Ni) as a source a liquid phase, and (3) proper selection of reactive tapes.

#### 14 of 16

# 4. Conclusions

Lightweight Ti/Ta/ceramics and Ta/ceramics layered composite platelets 1.5–2.5 mm thick can be prepared via a short-term (around 0.2–0.6 s) combustion of preliminary sandwich-like stacks of metal foils and reactive tapes. The structural/phase evolution and mechanical properties of synthesized layered composites are summarized as follows:

- Multilayer composites materials can be prepared via combustion reaction of reactive tapes that act as (1) a source of material for a ceramic transition layer and (2) a heatgenerating agent. The good joining of Ta with Ti was achieved due to SHS reactions in reactive tapes yielding ceramics and reaction heat.
- Layered composite material is formed as a result of mutual impregnation, interdiffusion, and chemical reactions. In contrast to the technique of ceramic–metal diffusion bonding, our combustion-aided process affords for joining dissimilar materials in short processing time without using complicated facilities.
- Good joining between metals and ceramics is reached due to the formation of interfacial layers in the form of cermets and eutectic solutions. Practically, all pairs—reactive Ti + 1.7B, Ti + 0.65C, 5Ti + 3Si tapes and Ti, Ta, Ni foils—were found capable of forming strong joints.
- Synthesized multilayered materials show good thermomechanical behavior.

Our approach can be recommended for use in designing materials with a preset structure, porosity, and mechanical properties.

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