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Investigation of the effect of compacting mode parameters of a carbon-aluminum wire preform on the strength of the produced compact

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The effect of the modes of compaction of a carbon-aluminum composite of a wire preform on its microstructure, properties, and nature of fracture has been studied. It is shown that with an increase in the temperature, load, and holding time, the volume fraction of pores decreases, and the fraction of fiber increases, which leads to an increase in the composite strength. In all cases, the shape of the fracture surface indicates the brittle nature of the composite fracture.

Keywords: MMC, carbon fiber/aluminum matrix composite, CF/Al-wire.

1. Introduction

One of the important problems of modern materials science is obtaining lighter and stronger materials. Pure metals and alloys have already reached the limit of their applicability; therefore, composite materials have assumed the role of new materials [1]. These include metal matrix composites reinforced with continuous carbon fibers [2]. Aluminum is one of the most relevant elements used as a matrix material due to its low density and manufacturability [3].

However, the production technology of carbon-aluminum composites has two significant problems: 1) carbon fiber is not wetted by aluminum melt; 2) a chemical reaction occurs at the carbon fiber/aluminum interface with the formation of hygroscopic aluminum carbides.

To solve these problems, various technological methods for producing carbon aluminum composite are being developed. They can be divided into two categories: liquid-phase and solid-phase. Liquid-phase methods include injection molding [4, 5, 9,10,11], mechanical introduction of fiber [6, 8], and ultrasonic treatment through the melt [7]. Below is a brief overview of the mentioned methods.

In [4, 5], a carbon-aluminum composite was produced by gas injection molding. The essence of the method consists in feeding liquid aluminum into a preform with carbon fibers and subsequent treatment due to gas pressure. The strength of the produced materials was 370 and 840 MPa, respectively. In [10], a composite was produced by injection molding followed by air cooling. The strength of the resulting composite did not exceed 300 MPa. As the authors reported, the low strength of the composite could be explained by the formation of aluminum carbides. A similar conclusion was reached in [11], where it was shown that with an increase in the melt temperature during injection molding, the composite strength monotonically decreased due to the formation of aluminum carbides. In [5], a composite was produced by a similar method. The composite had a fiber volume fraction of 30% and a tensile strength of 800 MPa. It should be noted that in cases of reaching a composite strength of 800 MPa and above, a nickel coating of the carbon fiber was used. According to the author, this explains a noticeably higher strength of the composite since, most likely, nickel fiber coatings slow down the reaction of aluminum carbide formation. There is also a method of liquid-phase stamping [8], which differs from injection molding in that it uses mechanical pressure rather than gas pressure. The composite produced by this method had a strength of about 490 MPa.

In [6], a composite was produced by the mechanical introduction of short fiber into the aluminum melt followed by mixing. The tensile strength of the resulting composite varied from 300 to 400 MPa. In [8], the strength of the composite produced by a similar method ranged from 86 to 172 MPa. Apparently, several factors affected the composite strength: uneven distribution of the fibers along the sample, the "sticking" of the fibers together, and the chaotic orientation of the fiber in the matrix.

In [7], the production of a composite wire by broaching carbon fiber through the aluminum melt subjected to ultrasonic treatment was reported. The strength of the resulting composite wire ranged from 400 to 1200 MPa, depending on the matrix alloy composition. Despite the sufficiently high strength, this method has significant limitations since it does not allow producing bulk material.

There is a method employing a special flux that provides wetting of the fiber with the matrix melt [9]: the fiber is pretreated with salt solutions (flux); after that, it is passed through the aluminum melt. It should be noted that the possibility of using this method as the technological one is

questionable since in the mentioned work [9] there are no data on the mechanical properties of the composite.

Most liquid-phase methods solve the problem of non-wetting of carbon fiber by creating pressure in the matrix melt, which compensates for the lack of wetting and the treatment of the fiber. However, the mentioned methods almost do not solve the problem of aluminum carbide formation, as pointed out by the authors themselves. Moreover, additional problems arise during composite production. One of them is the "sticking" of the fibers together and the unevenness of their distribution in the composite bulk. Another problem arises when moving from small samples to large-sized products, which will undoubtedly require the creation of enormous pressure to ensure complete treatment of the fiber. There is also a problem characteristic of the technology of large products. It is that the outer layers of the treatment preform are in contact with the matrix melt for more time than its core, which can lead to the deterioration of the mechanical properties of the product surface.

Solid-phase methods of production include those of powder metallurgy [10], layer-by-layer pressing of foils, hot vacuum pressing [8], and extrusion [11]. In powder metallurgy methods, short fibers are mixed with aluminum powder; after that, sintering is performed when exposed to temperature and pressure. When using extrusion methods, a mixture of powder and short fibers is extruded at an elevated temperature, resulting in compact sample formation. Note that the feature of solid-phase methods is the absence of a liquid phase, which allows one to solve both of the above problems. However, the main disadvantage of this type of methods is the use of short fibers. This significantly limits the volume content of the fiber, which in turn does not allow the maximum strength of the composite to be achieved. In addition, composites produced by a solid-phase method usually are isotropic. This means that when using such a material, there is no possibility of designing an optimal composite structure adapted to the stress state of the structural element [3].

In the present work, a combined method of producing a composite was employed, which consisted in producing a composite wire by liquid-phase treatment of a carbon fiber filament at the first stage and further compaction of the resulting wires at the second stage. The advantage of this method is the separation of the processes of fiber treatment and the formation of bulk material. This enables controlling the time of contact of the fiber with the melt with greater accuracy, and, therefore, controlling the formation of aluminum carbides. In addition, this method is very promising from the viewpoint of manufacturing large-sized products since it lacks the disadvantages inherent in other liquid-phase methods.

The work aimed to study the effect of compaction parameters on the microstructure, fracture pattern, and strength of the composite.

2. Materials and methods

2.1. Materials of study

UMT 40-3 K carbon fiber was used as a reinforcing element of the composite (manufactured by UMATEX Group, Moscow, Russia). The carbon fiber thread contained 3000 filaments, and the diameter of each filament was 7 μm. According to the manufacturer, the tensile strength and tensile modulus of the fiber were 4 and 260 GPa, respectively. Commercially pure aluminum (99.3 wt.%) was used as the matrix material.

2.2. Composite manufacturing

The composite was prepared in two stages. In the first stage, a composite wire was obtained by pulling the fiber through the aluminum melt in a similar way to how it was done in [17]. The production scheme is shown in Fig. 1.

From coil (1), carbon fiber (2) was pulled through the spinneret in the wall of graphite crucible (3) into the aluminum melt. The crucible was heated by induction heating (4). To ensure impregnation, the melt was subjected to ultrasonic treatment through niobium waveguide (5). After passing through the outlet die in the crucible wall, the wire was wound on coil (7). A detailed study of the effect of the parameters of composite wire production was carried out in [17]. In that work, it was shown that the melt temperature, broaching time, and ultrasonic processing power were of the greatest importance. In the present work, the melt temperature, the broaching time, and the ultrasonic treatment power had values of 700°C, 20 s, and 2 W/mm^2, respectively. The diameter of the wire produced through the melt was 0.4 mm.

In the second stage, the resulting wire was cut into 50-mm-long pieces and then laid into a preform for further compaction in a vacuum furnace. After the compaction, the composite samples had the following dimensions: a length of 50 mm, a width of 15 mm, and a height of 3 mm. The scheme of compact production is depicted in Fig. 2.

The effect of the compaction time was investigated in the range from 1 to 25 minutes with a step of 5 minutes at constant values of load and temperature of 1000 N and 700°C, respectively. The effect of the compaction temperature was studied in the range from 680 to 720°C with a step of

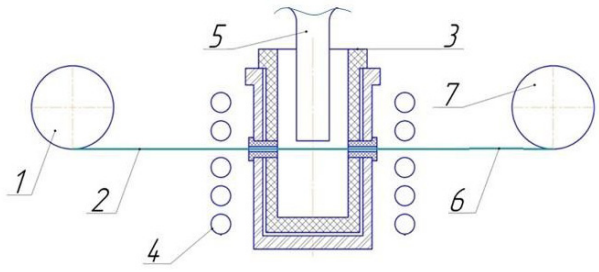


Fig. 1. Scheme of CF/Al wire production [17].

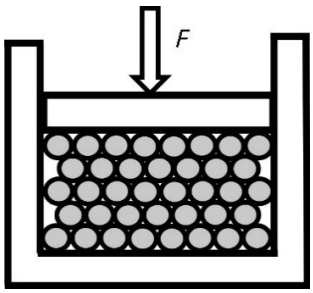


Fig. 2. Scheme of CF/Al composite production.

10°C at constant values of load and holding time of 1000 N and 1 minute, respectively. The effect of the compaction load was studied at 0, 100, 200, 500, and 1000 N at constant values of temperature and holding time of 700°C and 1 minute, respectively.

2.3. Research methods

The microstructure and fracture surfaces of the resulting composite were studied using a SUPRA 50 VP high-resolution scanning electron microscope. The images were taken in the secondary electron mode with accelerating voltages of 5 and 10 kV for the microstructure and fracture surfaces, respectively, at magnifications up to 10 000×.

The strength of the composite at three-point bending was determined by GOST 56805-2015. The test specimens had a width and height of 15 and 3 mm, respectively. The distance between the supports was 40 mm. The loading speed was 5 mm/min.

3. Results

3.1. Strength of the CF/Al composite

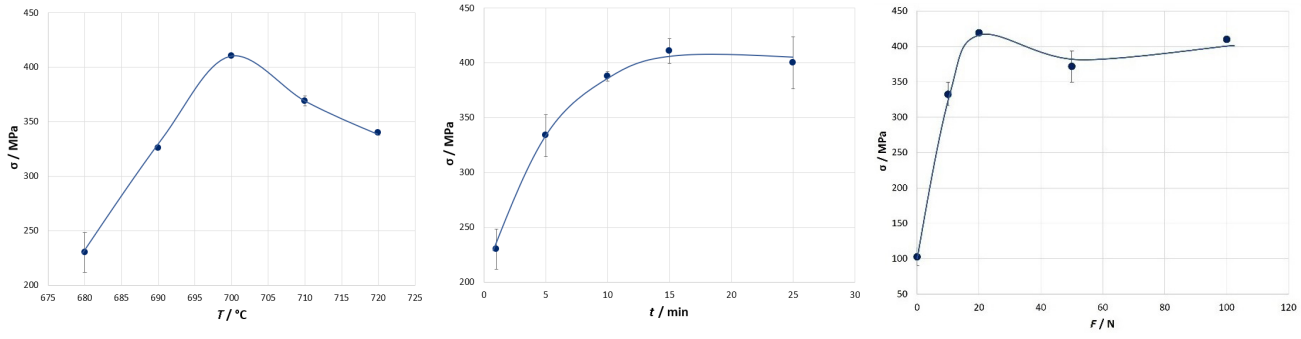
The dependencies of the strength of the CF/AL composite at three-point bending on the compaction parameters are shown in Fig. 3. With an increase in the holding time from 1 to 10 minutes, the strength increased from 230 MPa to ≈400 MPa (Fig. 3 a). With a further increase in the holding time, the strength value remained almost unchanged.

Depending on the compaction temperature, the strength changed from 230 MPa at 680°C to 400 MPa at 700°C (Fig. 3 b). With a further increase in the compaction temperature, a decrease in the strength to 350 MPa at 720°C was observed.

At zero load, the strength was 100 MPa (Fig. 3 c). With an increase in the load up to 200 N, the strength increased to 420 MPa. With a further increase in the load, the strength stabilized at about 400 MPa.

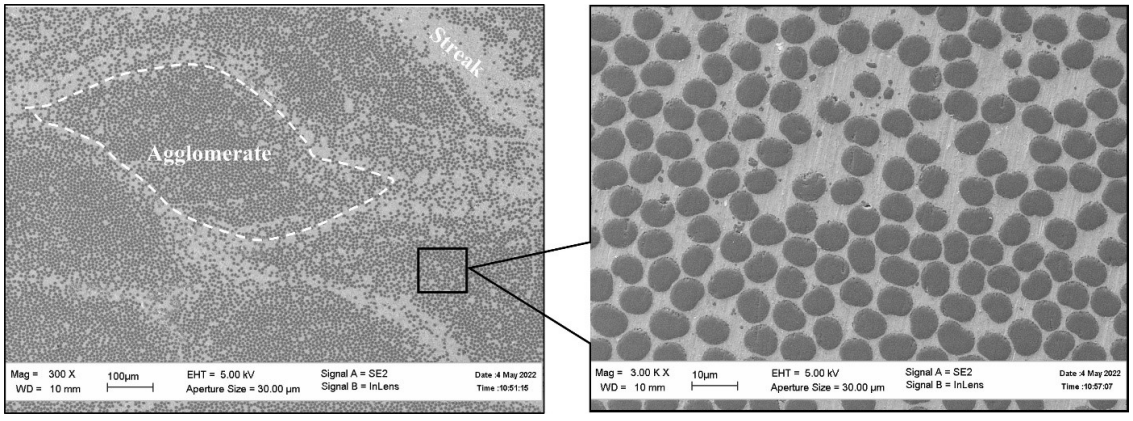
3.2. Microstructure of the CF/AL composite

The composite has a hierarchical structure that consists of two levels. At the first level, the structure is represented by agglomerates of carbon fiber filaments in an aluminum matrix. Between the agglomerates, there is a streak in the form of a matrix that is almost free of fiber, as shown in Fig. 4 a. Obviously, each agglomerate is formed by an individual composite wire. In this case, the interlayers between the agglomerates appear as a result of the sublimation of individual wires into a compact composite. The agglomerate structure, the second level of the hierarchical structure, is represented by individual carbon fiber filaments that are randomly arranged in an aluminum matrix, as illustrated



a b c

Fig. 3. Dependencies of the strength of the CF/AL composite at three-point bending on the compacting temperature (a), holding time (b), and compacting load (c).



a b

Fig. 4. Hierarchical structure of CF/Al composite, the first level of the structure, fiber agglomerates (a); the second level of the structure, agglomerate structure (b).

in Fig. 4 b. Individual filaments have a darker contrast as compared to the light matrix.

Figure 5 shows images of the CF/AL composite microstructure at different temperatures. It can be seen that, at a compaction temperature of 680°C, pores are observed in the composite structure between the agglomerates (Fig. 5 a). An increase in the temperature to 690°C leads to the almost complete disappearance of the mentioned pores (Fig. 5 b). In addition, at the streak between the agglomerates, some regions of the matrix are observed that are almost free of fibers. With an increase in the compaction temperature, the fraction of the mentioned regions decreases, and, thereby, the total volume fraction of the fiber in the composite increases, reaching its maximum value at 720°C (Figs. 5 c, d, e). In this case, the fraction of the streak matrix is reduced due to the displacement of its melt into the gaps of the open mold during compaction.

The holding time and the load during compaction have a similar effect on the composite microstructure.

3.3. Fracture surfaces of the CF/AL composite

Figure 6 illustrates the fracture surfaces of the CF/AL composite. At 680°C and 690°C, the fracture surface looks like a relief (Figs. 6 a, b); along with this, the streak and the pores between them are distinguishable. At 700°C, the fracture surface becomes flatter. Increasing the temperature above 700°C (Figs. 6 c, d, e) does not lead to any noticeable changes. At the same time, it is obvious that crack propagation occurs almost in one plane without any clear signs of fiber pulling out, which is also demonstrated in Fig. 6 f obtained at higher magnification.

The holding time and the load during compaction have a similar effect on the fracture surface of the composite.

4. Discussion

The comparison of the data obtained in the study of the microstructure of the composite, the nature of its fracture, and strength demonstrates a general pattern. It shows that with a decrease in the fraction of pores between the agglomerate and the fraction of the matrix streak (Figs. 5 a, b, 6 a, b), the composite strength increases to about 400 MPa (Figs. 3 a, b, c). This strength value is reached at the appropriate values of holding time, load, and temperature of 15 minutes, 1000 N, and 700°C. A further increase in the mentioned parameters does not lead to an increase in the composite strength. At the same time, the microstructure and nature of the fracture of the composite almost do not change.

Note that an increase in the temperature above 700°C leads to a decrease in the composite strength to 350 MPa. This is associated with the formation of aluminum carbides [18] that lead to the degradation of the fiber surface and, as a consequence, to a decrease in the strength of the composite itself. In addition, the formation of aluminum carbide leads to the formation of a strong interface between the matrix and the fiber [19].

The fracture surfaces of the composite, which has a strength of about 400 MPa, explicitly indicate that the composite fracture is brittle, i. e., after the initiation, the crack propagates in almost one plane. The load between the fibers is distributed unevenly and has a local character (LLS) [20]. There is a concentration of stresses on the fibers near the crack edge. While the fibers are sufficiently spaced from the crack edge, they almost do not perceive the load. Most likely, the observed character of fracture takes place due to the formation of a "strong" interface between the matrix and the fiber [21]. This is also indicated by the relatively low strength of the resulting composite.

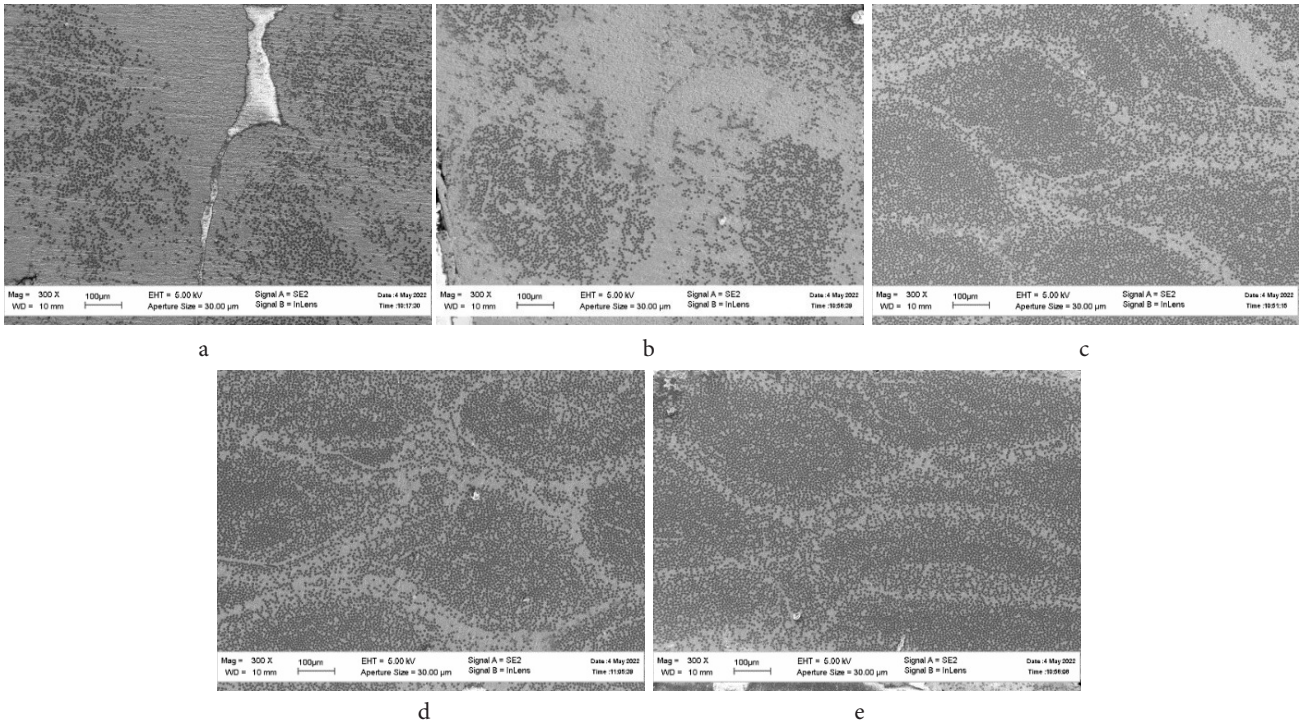


Fig. 5. Microstructure of the CF/AL composite depending on the compaction temperature of 680°C (a), 690°C (b), 700°C (c), 710°C (d), 720°C (e), at constant values of load of 1000 N and holding time of 1 min.

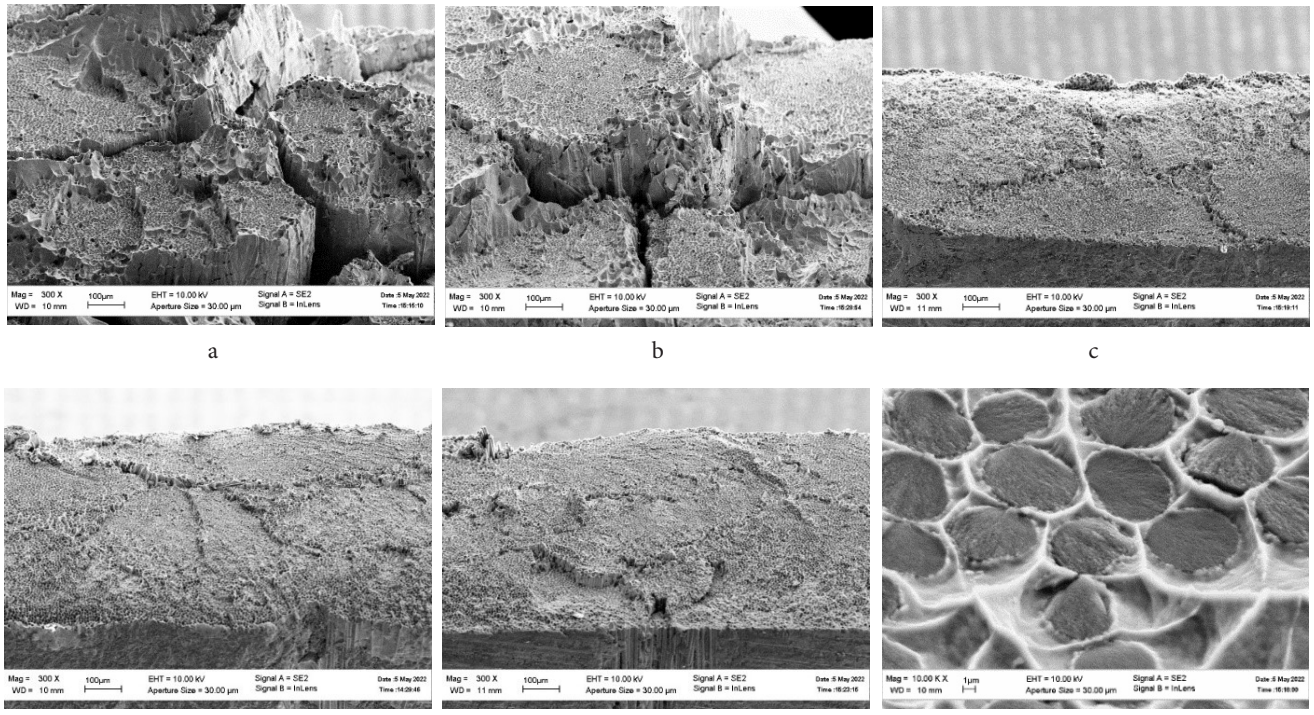


Fig. 6. Fracture surfaces of the CF/AL composite depending on the compaction temperature of 680°C (a), 690°C (b), 700°C (c), 710°C (d), 720°C (e), magnification of ×10 000 at 700°C (f), at constant values of load of 1000 N and holding time of 1 min.

The above review of recent results demonstrates the obvious disadvantages of the methods used to produce a composite. At the same time, in most cases, the strength of the composite produced in this work exceeds that of the analogs produced by other methods. This indicates the prospects of the proposed method, especially for producing large-sized products, which was shown in the Introduction.

Nevertheless, the strength level of the composite produced by the proposed method remains low, does not exceed the most durable aluminum alloys, and is significantly lower than that of carbon fiber composites. This implies that the practical application of the proposed method is possible only in combination with techniques that provide sufficient strength, for example, with the preliminary application of barrier coatings onto carbon fiber.

5. Conclusions

1. A two-stage method of producing a carbon-aluminum composite has been proposed, which consists in producing a composite wire by liquid-phase treatment of a carbon fiber filament at the first stage and further compaction of the produced wires at the second stage. The produced composite has a hierarchical structure. At the first level, the structure is represented by agglomerates of carbon fiber filaments in an aluminum matrix. Between the agglomerates, there is a streak of the matrix that is almost free of fiber. The agglomerate structure, the second level, is represented by carbon fiber filaments that are evenly arranged in an aluminum matrix.

2. With a decrease in the fraction of pores between the agglomerates of the preform and the fraction of the matrix streak, the composite strength increases to approximately 400 MPa. This value of strength is reached at the appropriate values of holding time, load, and temperature of 15 minutes, 1000 N, and 700°C. A further increase in the mentioned parameters does not lead to an increase in the composite strength. At the same time, the microstructure and nature of the fracture of the composite almost do not change.

3. The destruction of the composite of the highest strength (400 MPa) is brittle. Crack propagation occurs almost in the same plane, which is most likely due to the formation of a "strong" interface between the matrix and the fiber, which is also indicated by the relatively low strength of the resulting composite. This implies that the practical application of the proposed method is possible only in combination with techniques that provide sufficient strength, for example, with the preliminary application of barrier coatings onto carbon fiber.

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