# WEAR APPLICATION OF Co-TiC CEMENTED CARBIDES FOR GAS TURBINES

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## Summary

The work offers a solution of urgent theoretical and practical problem of increase of wear resistance of GTE blades top shrouds contact faces. Implementation of current research results may increase service life of turbines.

For Co–TiC cemented carbides manufacturing TiC-powders of grain size -20…+0.5 µm were used. This provided them with combined micro and macro hardening. High strength of an alloy helps it to resist plastic deformations arising due to action of friction force. Accompanied by generating of friction induced surface structure with positive gradient of mechanical properties provides to Co-TiC cemented carbides perfect wear resistance at elevated and high temperature operation

INTRODUCTION. Developments in aerospace engineering were always accompanied by increase of fuel combustion temperature, thus improving fuel efficiency or increasing thrust of the engine (ref. 1). But it also increases loading of turbine elements, especially of rotor blades and decreases service life of the engine by causing more rapid loss of surface material due to wear (ref. 2). Particularly blades top shrouds contact faces should be mentioned. The point is that nickel-based alloys, used to produce blades, have poor wear resistance. Plastic deformations of material surface layer due to friction cause a rise of oxidation and fatigue wear.

To solve the problem application of Co-TiC cemented carbides to cover top shrouds contact faces is proposed. They have higher melting point (1320–1340°C) then majority of used today nickel-based alloys. Their high temperature creep resistance is provided by carbide fraction +0.5…1 µm by a mechanism, similar to $\gamma'$ precipitate at nickel alloys. Carbide grains -20…+5 µm withstand wear thus reducing stresses in matrix material and preventing surface layer plastic deformations.

Current work is devoted to experimental investigation of their high temperature fretting wear resistance.

EXPERIMENTAL FACILITIES AND MATERIALS. For experiment ring-to-plane wear test with ring reciprocal rotating motion was used. Applied relative displacement – 120 µm, frequency – 30$s^{-1}$, number of displacements per test – $5\times10^6$, specific load – 30 MPa. Temperature ranged from 300 °C to 1050 °C, what was provided by circular electric furnace. Applied relative displacement, specific load temperature values are close to operating conditions of engine D-36. Specimens of powder alloys for wear test where produced by hot-vacuum-sintering process. Cutting them on ready-to-soldering parts was performed by spark cutting in oil bath. After preparation, they where jointed to the holders by high temperature solder with melting point 1240–1270°C (ref. 3). The view of soldered to holders specimens is given on fig. 1.

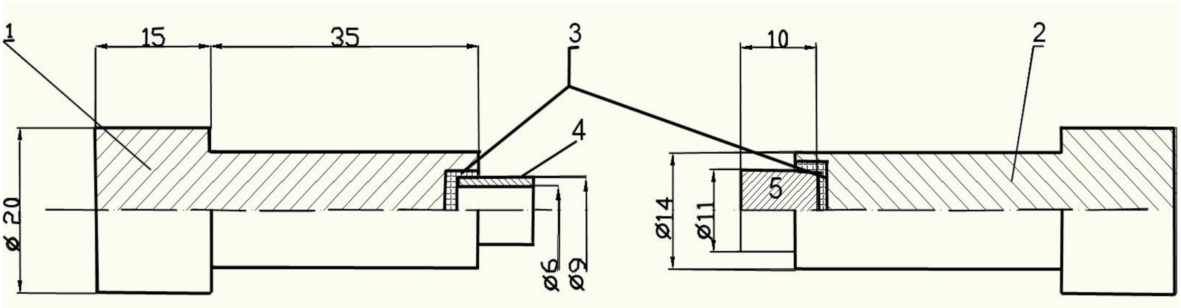
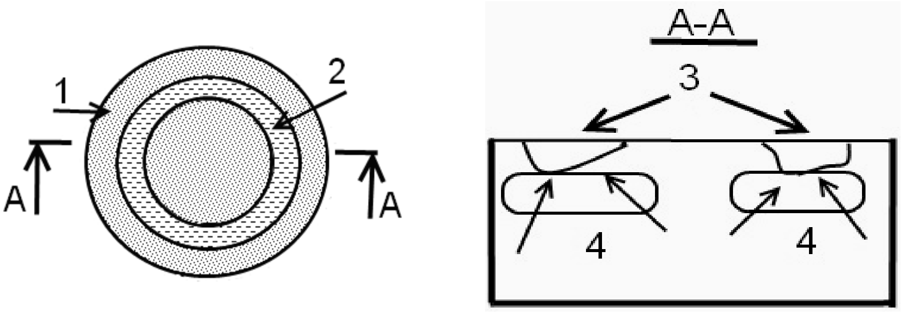


Fig. 1. Diagrammatic representation of specimens: 1, 2 – holders, 3 – solder, 4 – movable specimen, 5 – fixed specimen

TiC powders for sintering where prepared by high-energy ball milling with separation of +20 μm grains. Titanium carbides in quantity 50 % vol. (alloy P-76) have been cemented by cobalt. To increase high temperature properties of the binder it was additionally alloyed by 20–24 % (mas.) of chromium and by 3–4 % (mas.) of aluminium and iron. Wear test results were compared with Co-TiC cast alloy containing 30% vol. of carbides (alloy P-69 (ref. 4)) and industrial eutectic alloy XTH-62 (ref. 5), which contains up to 20 % vol. of NbC. To determine wear resistance linear wear of specimens was measured (ref. 6). After wear test specimens were detached from holders and their cross-sections were made in order to determine plastic deformation of sublayer (fig. 2).



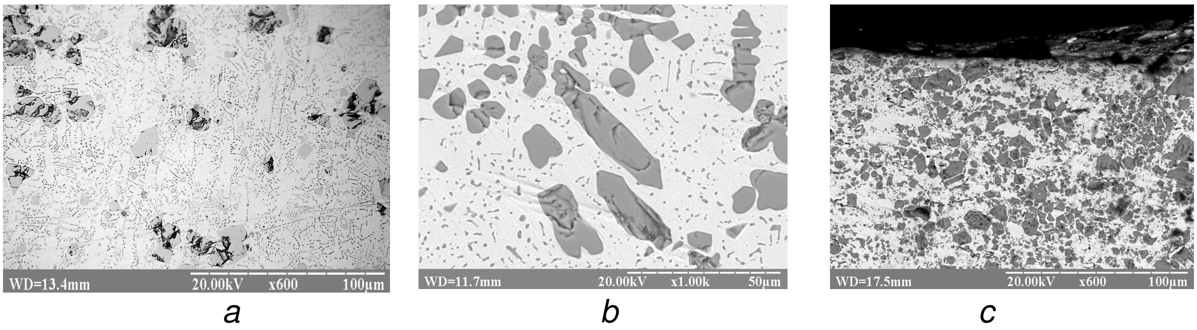
*a* *b*

Fig. 2. Specimen top (a) and section (b) views: friction surface (1), wear scars top (2) and cross-section (3), surface deformation analyzed area (4)

STRUCTURAL ASPECT. Both cast alloys structure is obtained due to crystallization of carbides solution in cobalt. As far as titanium (niobium) and carbon are introduced as separate phases, it takes some time to full reaction running. The primary carbides solidification starts much earlier then that of binder. This process runs until the level of solubility of carbides in liquid phase. This leads to significant growth of their size (up to 30 μm) and irregular distribution through the alloy. The cracks in these grains are clearly seen (fig. 3, *a*). As volume ratio of carbides in binder increases, TiC and NbC grain growth up to 60μm (fig. 3, *b*). It is also necessary to mention, that 30% vol. of carbides in eutectic alloys is limiting and require elaborate manufacturing procedures due to high viscosity of a melt.

During binder solidification smaller (secondary) grains of 0.5–2 µm size appear. Their quantity is predetermined by solubility level of carbides in molten cobalt and can hardly be changed significantly without additional heat treatment.

Powder alloy P-76 (ref. 4) (fig. 3, *c*) is characterized by uniform carbides distribution through the alloy. Grain size varies from 0.5 µm up to 20 µm.



*a* *b* *c*

Fig. 3. Microstructure of cast alloys XTH-62 (*a*), P-69 (*b*) and cemented Co-TiC carbides (*c*)

Pore size seldom rises to critical (6–12 µm) (ref. 7), it's density is as high as 97% of theoretical. Pores in powder materials cause failure initiation and are strongly unwanted (ref. 8).

EXPERIMENTAL RESULTS. Wear test showed an advanced wear resistance of Co-TiC cemented carbides. Their linear wear is much less then of other tested materials what can be compared using fig. 4.

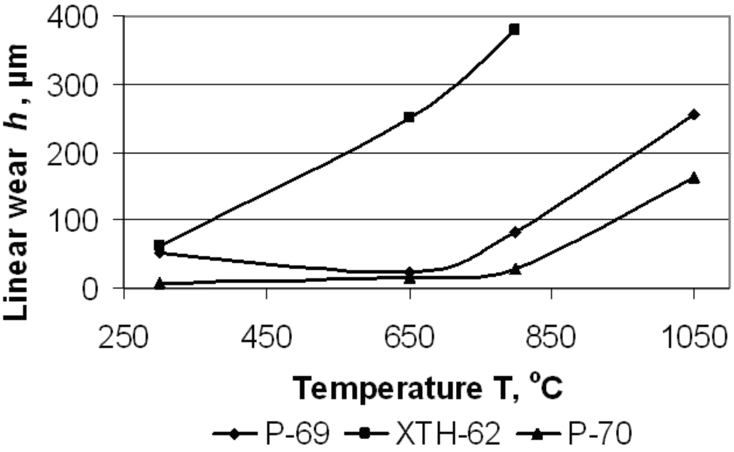
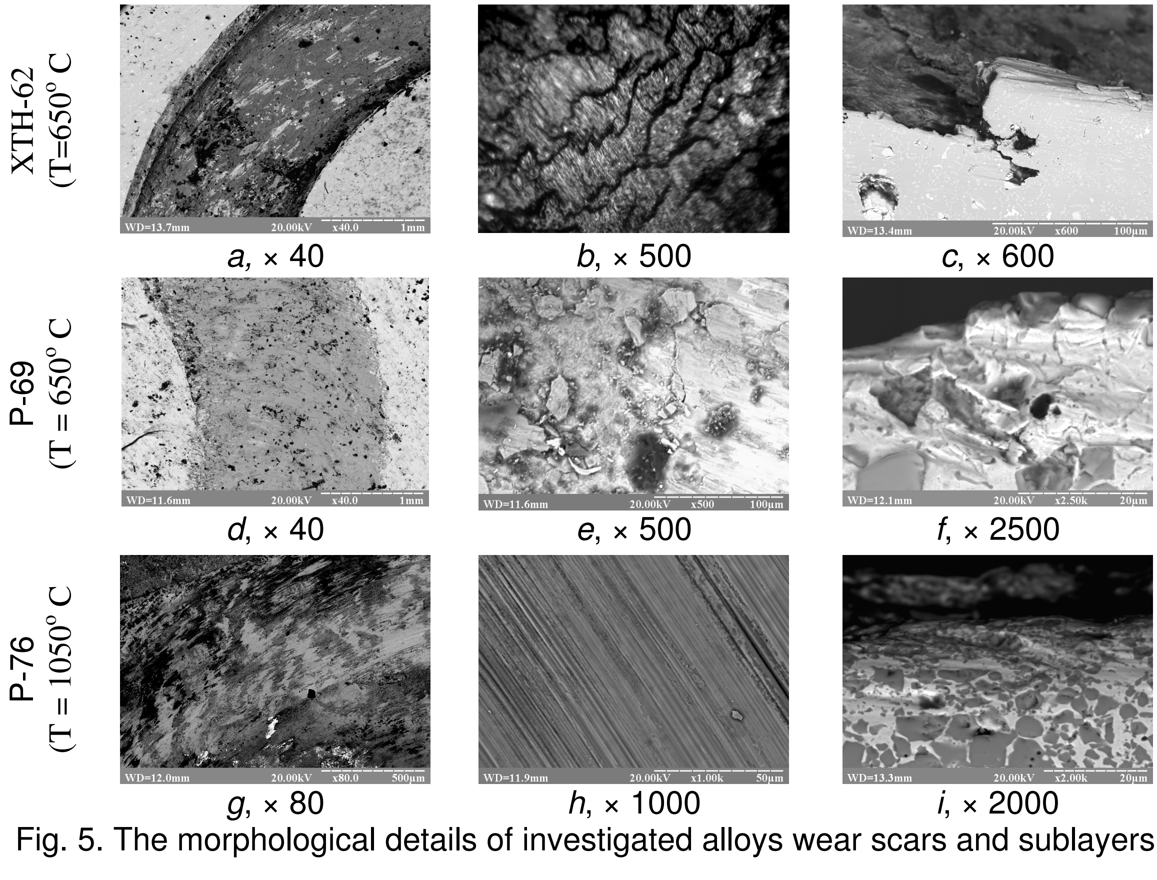


Fig. 4. Wear chart of investigated materials

Top view of wear scars analyses gives a conviction in wear of XTH-62 alloy due to intensive oxidation (fretting-corrosion). On its friction surfaces thick and porous oxide layer is formed. It is destructed on actual contact areas, as a result contact of two unprotected by oxide layers metallic surfaces becomes possible (bright areas on fig. 5, *a*). Analyses of bright areas on wear scars helps to make conclusion about their formation as a result of fatigue wear. It is proved by surface "terracing" (fig. 5, *b*) which is typical for this kind of wear. High loads due to friction cause plastic deformations of sublayer (fig. 5, *c*). Taking into account these two simultaneous processes it can be deduced that friction-caused plastic deformation rapidly increases chemical activation of cobalt-based solid solution resulting in high oxidation and fatigue wear rate.

Similar behavior was observed at analyzing of P-69 alloy. Despite lower wear it's surface is also deformed. It can be clearly seen on fig. 5, *d*. Plastic deformation causes high strain in surface layer. This leads to destruction of friction-induced surface oxide films (fig. 5, *e*). It can be explained both by high external loading and plastic deformations of sublayer. The letter is proved by fig. 5, *f*.

Results of P-76 powder alloy testing differ a lot from eutectic. At high temperature it forms a discrete surface structure. It represents a mixture of bright and dark fields (fig. 5, *g*). Size of the letter changes from 100 up to 300 μm. Friction-induced surface structures were analyzed for chemical composition using scanning-probe microscope REMMA 106N. It has proved that different fields (dark and bright) are being formed by different oxides: bright fields contain up to 50% (mass) of chromium and 4–6% (mass) of aluminum, dark fields contain up to 46% (mass) of titanium and almost free of chromium and cobalt. Material of both fields contain about 40–48 % (mass) of oxygen. By this it may be proved, that dark fields are formed by titanium oxides on areas of continuous friction contact, what is partially proved by their dimensions. Bright fields are formed on "friction-free" areas. On microlevel friction-induced film is smooth, with no signs of fatigue wear or cracks (fig 5, *h*). Cross-section of specimens under wear scars (fig. 5, *i*) proves absence of sublayer plastic deformation.



*a,* × 40 *b*, × 500 *c*, × 600

*d*, × 40 *e*, × 500 *f*, × 2500

*g*, × 80 *h*, × 1000 *i*, × 2000

Fig. 5. The morphological details of investigated alloys wear scars and sublayers

DISCUSSION. Plastic deformation of sublayer is important factor which causes intensive fatigue and oxidation wear. Also heated metal is "extruded" (fig. 5, *c*) from the most loaded contact areas. Thus, to increase wear resistance of material it is necessary to avoid plastic deformations of sublayer. Use of 50 % (vol.) of TiC is efficient to solve this task. High temperature causes oxidation of both binder and filler. During this process fine and submicron $TiO_2$ worn particles are formed (fig. 6). High temperature (up to 1050 $^{o}$C), small size of particles activate grain boundary diffusion, and high pressure on friction areas leads to their sintering due to capillary forces (ref. 9) over binder oxide layer thus forming friction induced structure with positive gradient of mechanical properties. Even if not sintered, $TiO_2$ particles are not removed from contact area during fretting-wear test. Their hardness is less then hardness of binder oxide layer and they may perform a function of solid lubricant (ref. 10).

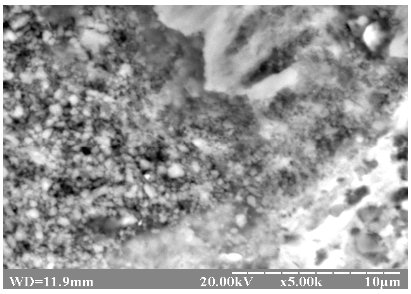


Fig. 6. Formation of fine and submicron $TiO_2$ worn particles

CONCLUSIONS. Use of powder metallurgy methods allows controlling of carbides grain size and their distribution through the alloy. Precisely balanced volume ratio of carbides makes it possible to reduce plastic deformations and to decrease wear of Co-TiC cemented carbides.

Oxidation of TiC surface leads to appearance of fine and ultra-fine $TiO_2$ powder. Simultaneous action of high temperature and pressure in contact area make them sintered to the matrix oxide layer, thus gradient friction-induced structure is formed. Results of wear tests performed in close-to-operating conditions showed an advanced wear resistance of Co-TiC cemented carbides. It makes them prospective for gas turbines wear applications.

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