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Vacuum Annealing Effect on the Wear Resistance of Tic-Tin-Al_2O_3 Cutting Tool Nanocomposites

Zviad Kovziridze[1], Gulnazi Tabatadze[2], Tamar Loladze[3], Natela Nizharadze[4]

[1,2,3,4] Georgian Technical University, Institute of Bionanoceramic and Nanocomposite Technology, Faculty of Chemical Technology and Metallurgy. 69, Kostava Str., Tbilisi 0175, Georgia

ABSTRACT:

Objective: The aim of this study is to improve the physical and mechanical properties of the TiC-TiN-Al_2O_3 system nanocomposite [1,2] developed at Georgian Technical University.

Method: The effect of vacuum annealing at various temperatures and durations on the cutting and wear resistance properties of the material was studied. The cutting properties were examined on a lathe, and wear resistance was determined using microscopic methods.

Results: It was found that the mechanical properties of the nanocomposite slightly improved under the influence of annealing temperature and duration. In terms of wear resistance and cutting properties, improvements of 25-30% were observed. The optimal annealing parameters are as follows: temperature - 600°C, time - 60 minutes, vacuum pressure - 10^{-2} Pa.

Conclusion: To enhance the wear resistance and cutting properties of the TiC-TiN-Al_2O_3 system nanocomposite, a study was conducted to investigate the effects of vacuum annealing at 1400°C and 25 MPa pressure. The results showed that annealing reduced the surface defects and residual stresses in the material, leading to significant improvements in wear resistance and cutting properties. Specifically, the wear width of the cutting edge on the tool's back surface decreased from 0.43 mm to 0.29 mm, and the tool's operational lifespan increased by 10-25 minutes, depending on the cutting speed.

MAIN PART

At Georgian Technical University, under the leadership of Professor Teimuraz Loladze, numerous studies have been conducted to develop advanced instrumental materials and investigate their properties [3-7]. Under the leadership of Professor Zviad Kovziridze, research in this area is progressing successfully, with a focus on developing new compositions with enhanced technical properties and refining existing ones [8-26].

Based on these studies, TiC-TiN-Al_2O_3 composite was developed [1,2], with the optimal composition presented in Table 1. The research utilized Al_2O_3 nanoparticles (435 nm) from Germany's ALCOA company, along with TiC and TiN, which have specific surface areas of 4 m²/g and 5 m²/g, respectively. The crystal lattice parameters of the components in the initial composite mixture are as follows: TiC - d_{hkl} - 2,495; 2,160A; TiN-2,445, 2,117A, Al_2O_3- d_{hkl} - 3,476; 2,555A; The composite was produced via hot pressing at a temperature of 1400°C. During the TiC-TiN-Al_2O_3 sintering process, aluminum oxide does not react with any other components, while TiC and TiN form a titanium carbonitride structure. The process of titanium carbonitride formation was studied (Figure 1), where the starting materials were titanium carbide and titanium nitride in the same ratio as used in the composite.

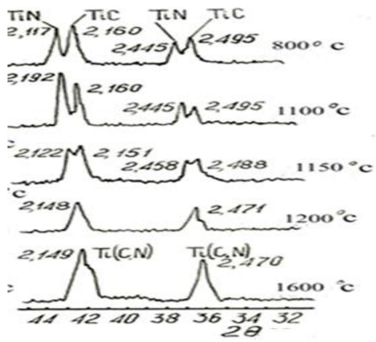

Fig.1. X-ray structural analysis of the transformation process of titanium carbide and titanium nitride into carbonitride

The analysis indicates that the formation of a titanium carbonitride solid solution begins at 1150°C, with a distinct titanium carbonitride reflection appearing at 1200°C. By 1600°C, the crystalline structure of titanium carbonitride is fully established. In the composite, well-defined carbonitride peaks are evident as early as at 1400°C, with d_{hkl} values of 2.274, 2.030, and 1.516. The aluminum oxide reflections remained almost unchanged, as depicted in Fig. 2.

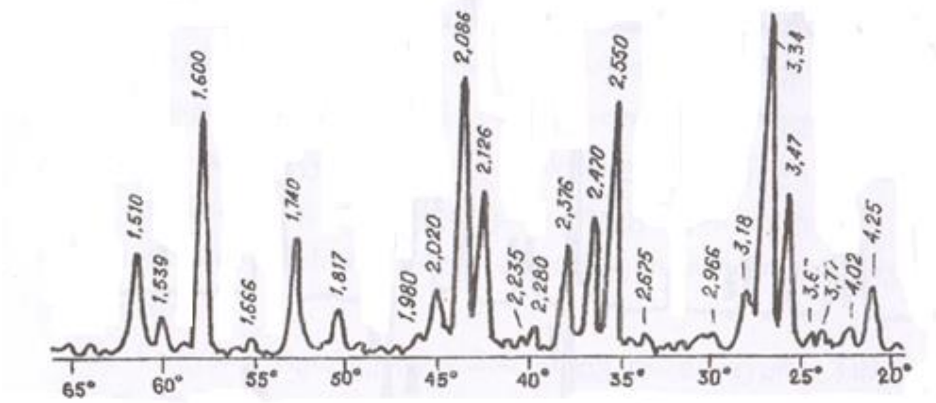

Fig.2. X-ray structural analysis of a composite obtained by hot pressing in the TiC-TiN-Al_2O_3 system

The microstructure of the resulting composite (Fig. 3) reveals that carbonitride grains are dispersed throughout the aluminum oxide matrix. The composite's physical and technical properties are detailed in Table 1.

Table 1: Composition, mass % and properties of the samples obtained by hot pressing of the TiC-TiN-Al_2O_3 system composite

Sintering temperature, T, ^{0}C	Al_2O_3 Med. 435 nM	TiC	TiN	MgO	Relative density, ρ g/cm^2	Micromechanics by Rockwell, HRA	Compression strength, Σp. MPa.	Bending strength, $σ_{b.}$ MPa	impact strengtha, kJ/M^2	The width of the wear edge on the rear surface of the tool, μM
1400	79	10	10	1	0,97	91,5	2450	570	21	430

Fig.3. Microstructure of the TiC-TiN-Al_2O_3 system, X1500.

The operational properties of any ceramic material or composite are largely determined by the characteristics of its main component, the crystalline phase, a concept reflected in Professor Zviad Kovziridze's formula. [27].

The dependence of the micro- and macro- mechanical characteristics of the materials on the crystalline phase content in the composite was calculated according to Z.Kovziridze's [27] formula:

$$\sigma_d = \frac{\text{P.Fkd}}{\text{Km Kv Fkf'}}$$

Where: P-load; Km- middle size of crystals; Kv- volume fraction of crystals in the matrix; F_{kd}-crystals distribution factor in the matrix, which is determined by the researcher; in case of equal distribution it equals to 1, in case of unequal distribution = 0.9; F_{kf}-crystal shape factor, is taken as the ratio of the largest characteristic size of a crystal to the smallest, which allows us to characterize the shape of a given set of crystals, according to which we are able to define correlation of mechanical characteristic in the matrix from the crystal phase characteristics in our proposed formula.

Table 2: Crystalline phase characterization and mechanics of TiC-TiN-Al_2O_3 nanocomposite system

Visual field, μM^2	Maximum grain size, $K_{max.}$,µM	Minimum grain size, $K_{min.}$ µM	Fraction of the crystalline phase in the composite, %	Grain shape factor, $F_{kf}=K_{max.}/K_{min.}$	Grain distribution factor, F_{kd}	Bending strength σ_b. MPa	Compression strength, σp. MPa
28000	12	2	96	6	0,9	570	2469

Based on these data, the σ_d values were calculated for both bending and compression.

For bending: σ_d=570·0,9/12·96·6 =0,25

For compression: σ_p=2469·0,9/12·96·6 =1,1

It is known [28] that the surface of a material typically exhibits inhomogeneity in its microstructure, as well as residual stresses and surface defects, which reduce the resistance of oxide-containing materials to brittle fracture. One approach to reducing surface defects and enhancing the reliability of cutting materials is the vacuum sintering method [29-31].

Material fracture during cutting occurs not through a single mechanism, but through a variety of distinct mechanisms, depending on the material and the conditions. The process is further complicated by the fact that the same material may fracture in different ways, depending on the load stress, loading rate, material's prior history, environmental conditions, and temperature. Ignoring any of these factors, or failing to account for the various fracture mechanisms, can lead to significant errors. For most ceramic materials, fracture occurs in a brittle manner, meaning they undergo little to no plastic deformation. In the cutting process, the interaction between the cutting tool and the workpiece is of paramount importance, given their properties and the complexity of the cutting process.

If an isotropic solid body undergoes elongation during the cutting process due to mechanical stress modes, the atomic distance must increase. The elongation of the sample corresponds to the expenditure of a certain amount of energy for its elastic deformation. This energy for elastic deformation should be expended in increasing the system's surface energy as new surfaces are formed during the cutting process. If the specific surface energy is equal to ***y***, then the energy required to form new surfaces during the cutting process is ***2y***. A significant portion of the energy required for cutting the sample must be accumulated between the atoms (or molecules) that are directly adjacent to the fracture (cutting) surface. Thus, σmidl. if is the stress at which cutting (or fracture) begins, and σmid.^{2}/2E represents the energy density of elastic deformation, then Hooke's law should be considered, and the value of the elastic modulus is typical. In this case, the energy σmid.^{2}/2E, which is distributed over two atomic planes separated by a distance ***a***, should be equal in magnitude to the surface energy ***y***. Accordingly, the energy required for cutting, σmid., can be determined from the equation:

$$\sigma_{mid.} \approx \sqrt{2yF/a} \approx E / 10$$

For crystalline oxides, whose elastic modulus falls within the range of (1-5) × 10^5 kg/cm², this corresponds to the theoretical strength [32]. The strength of industrial high-alumina ceramics is approximately 3500 kg/cm² (E/100), while the fracture strength of ordinary porcelain is significantly lower. The reason for this difference between theoretical and practical strength was first explained by A.A. Griffith [28]. Numerous studies have shown not only isotropic brittle materials containimg cracks that reduce their strength compared to the theoretical strength, but also that these cracks are located on the surface, not within the material. This is why thermal treatment, such as annealing, has a positive effect before the cutting process. In this case, the reduction of the material's internal volume during cooling after annealing creates compressive stresses in the relatively cooler surface layers. As a result, in order for the tensile stresses at the sample's surface to reach the strength limit and enable effective cutting, a relatively high load must be applied. Thermally treated cutting material will exhibit relatively higher tensile stresses. This demonstrates that the cutting process begins at the expense of the dangerous cracks present on the surface of the test material. This phenomenon also suggests that the strength of porcelain increases when a coating (e.g., glaze) with a lower thermal expansion coefficient is applied, leading to compressive stresses. Of course, the applied tensile load must be higher to overcome the surface stresses of the cutting material and initiate cutting.

When the cutting tool comes into contact with a solid material, dislocations are generated in the crystals of the cutting material. This interaction may lead to the formation of microscopic cracks on the surface, as significant stresses develop upon contact with the cutting material. Another source of micro-crack formation on polycrystalline cutting tools may be the difference in the coefficients of thermal expansion between the phases present, leading to increased stresses at the grain boundaries. Such stresses can be high enough to cause fine cracks and weaken the cutting material. Micro-cracks may also result from thermal stresses created during the cooling process after annealing of the cutting tool, or from thermal stresses generated during the cutting process, especially if there is high stress on the surface of the cutting tool and workpiece. The concentration of stresses and crack formation in multiphase materials is also influenced by the development of micro-channels at the surface between different phases or between grain boundaries in the same phase due to thermal poisoning. Ultimately, these processes are caused by mechanical damage to the surface, as well as its chemical and thermal corrosion. It should be noted that disregarding these processes is very difficult in practice. According to theory, fracture may occur because dislocations generated during plastic deformation can combine to form micro-cracks, which lead to the onset of brittle fracture. It should also be noted that brittle fracture can be observed not only in brittle materials but also in ductile materials. In such cases, it is observed that some plastic deformation always precedes fracture. Dislocations may cluster in large numbers at obstacles that impede their movement, such as the intersection of slip planes, grain boundaries, and surfaces. The concentration of dislocations at these sites creates high local stresses, which can be sufficient for several dislocations to combine and initiate crack formation. One example of this is shown in Fig. 4.

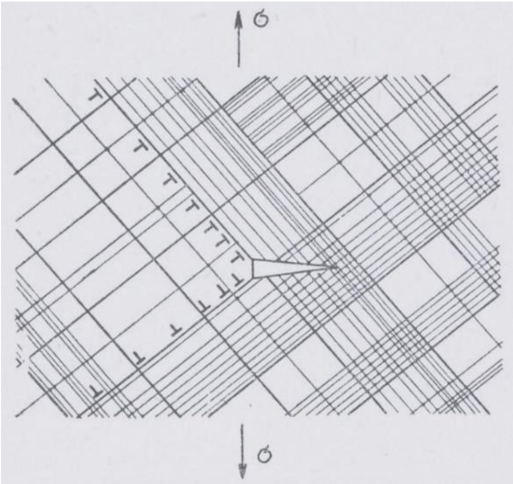

Fig.4. Schematic representation of dislocation aggregation and crack formation at the intersection of two slip lines

During the rapid generation of dislocations in the process of plastic deformation, increasing numbers of dislocations concentrate around pre-existing microcracks, leading to their gradual growth. If the movement of dislocations is impeded by impurities or if they are generated under high stress, microcracks can grow due to stress concentration at their tips, potentially causing further crack propagation, as commonly observed in brittle materials. In brittle materials, such as cutting tools, microcrack growth requires less energy compared to ductile materials. In ductile materials, plastic deformation around the crack tip delays further propagation, whereas in brittle materials like ceramics, crack growth proceeds more readily, often leading to catastrophic failure at the same stress levels. Therefore, for brittle materials, the critical stage in the fracture process is crack initiation.

A different case applies to carbides with metallic binders, which are used as cutting materials. In this case, during fracture progression, a new crack must be initiated at each boundary between phases. The plasticity of the metallic phase requires significant energy expenditure to allow crack propagation, and the stress needed for further crack growth must increase. The materials, such as cutting tools, do not fracture even when the applied stress exceeds the stress required for crack initiation.

In order to improve the properties of the composite system we developed (TiC-TiN-Al_2O_3) [1,2], tests were conducted using a vacuum annealing method (10^{-2} Pa), as shown in Table 1 and Figures 1-3.

For the experiment, we prepared samples using the hot pressing method at a maximum temperature of 1400°C with an 8-minute hold time under 25 MPa of pressure. The samples were divided into two groups: the properties of one group were measured both before and after vacuum annealing. Table 1 presents the chemical composition of the composite under investigation, the average results of its mechanical properties, and the wear rate. Figure 3 shows the microstructure of the composite, where carbonitride grains are distributed within the aluminum oxide matrix.

Vacuum annealing experiments were conducted in a vacuum furnace at a pressure of 10^{-2} Pa, varying both the sintering temperature and duration. Specifically, the temperature ranged from 500°C to 800°C, with 100-degree increments, while the annealing duration varied from 30 to 120 minutes, with 30-minute intervals. No significant effect of the annealing process on the microstructure characteristics was observed, but among the mechanical properties, the hardness increased from 91.5 to 92.0 HRA. Table 3 and Figure 5 present the results of the tool durability tests equipped with cutting inserts made from the nanocomposite after vacuum annealing.

Table 3: Dependence of the cutting edge width wear on the temperature and duration of annealing

Annealing temperature T ^{0}C	The width of the cutting edge wear on the back surface of the tool, μM			
	Duration of the annealing process, min.			
	30	60	90	120
500	400	380	380	380
600	320	290	280	280
700	310	280	290	280
800	310	270	280	280

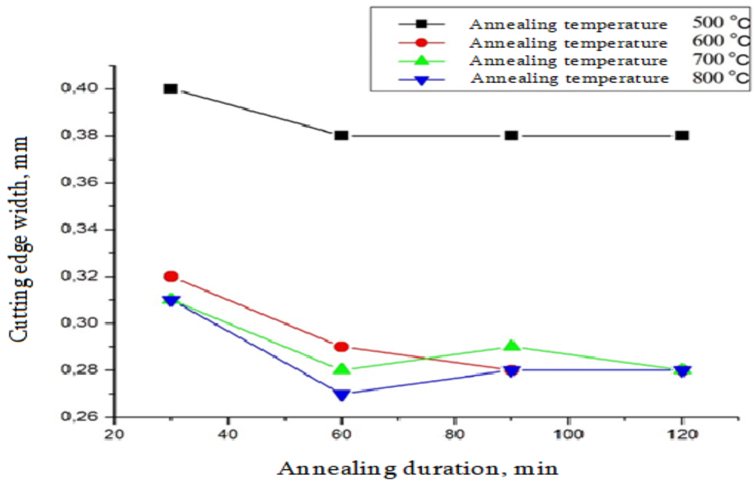

Fig.5. Curves of dependence of the cutting edge width wear on the temperature and duration of annealing

Durability tests were conducted on a Model 163 turning lathe, using steel grade 45 as the workpiece material. The width of the flank wear on the cutting tool was measured using a microscope.

As the experiment showed, the annealing temperature had a relatively greater effect on the wear rate compared to the sintering duration, which is in full agreement with the conclusions drawn in reference [28]. The optimal parameters for the annealing process were determined to be a temperature of 600°C and a duration of 60 minutes. As seen from the table, while increasing the temperature above 600°C slightly decreases the wear rate, the reduction is so minimal that, from an energy-efficiency standpoint, it is preferable to conduct the process at 600°C. When comparing the width of the tool wear before and after annealing, wear resistance increases by an average of 25-30%.

The cutting properties of the samples before and after annealing were tested on steel grade 45 during turning and on the titanium alloy BT3-1 during milling. The results are shown in Figures 2a and 2b.

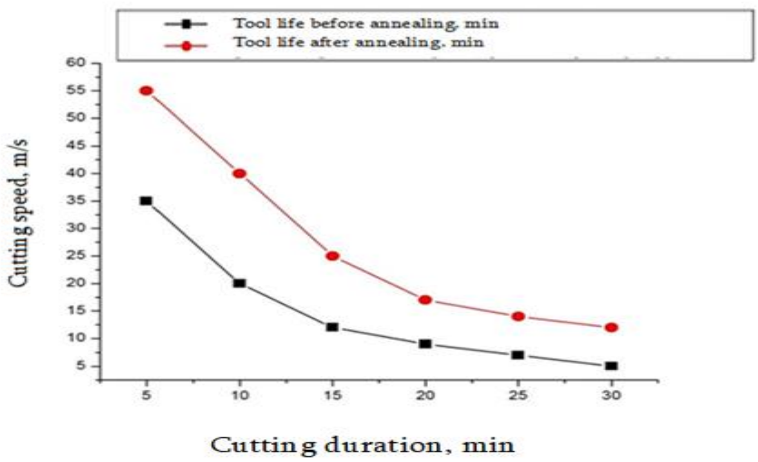

Fig.6. V-T dependence curves while turning the steel 45 on the lathe.

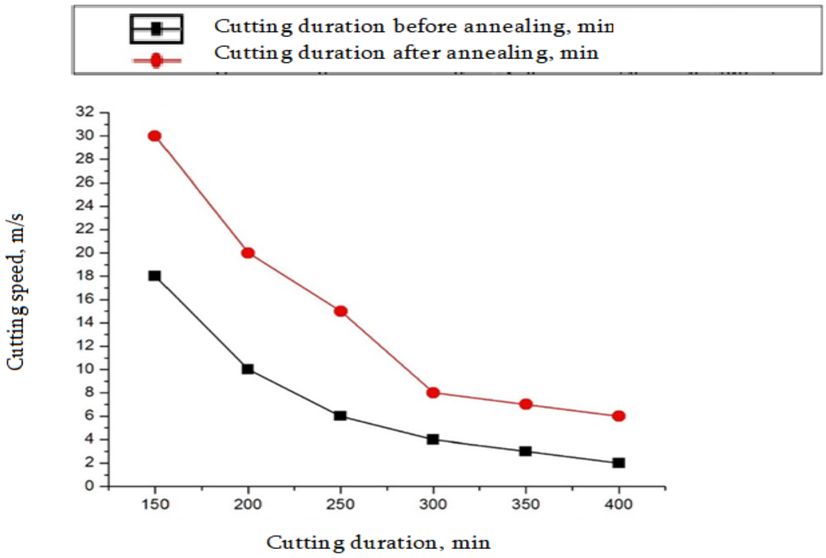

Fig.7. V-T dependence curves during milling of BT3-1 alloy

From the dependence of time and cutting speed, it is evident that in both cases, the cutting properties of the composite improved as a result of vacuum annealing, which leads to an increase in the tool reliability coefficient and productivity.

CONCLUSION

The optimal parameters for vacuum annealing of TiC-TiN-Al_2O_3 system nanocomposite samples, obtained by hot pressing at 1400°C, with an 8-minute hold time under 25 MPa of pressure, are as follows: temperature - 600°C, sintering duration - 60 minutes. The wear resistance and cutting properties of the samples obtained under these conditions were significantly improved (Tables 1 and 2).

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