INTERNATIONAL JOURNAL
MATERIALS SCIENCE. NON-EQUILIBRIUM PHASE TRANSFORMATIONS

PUBLISHER:
SCIENTIFIC TECHNICAL UNION OF MECHANICAL ENGINEERING
108, Rakovski Str., 1000 Sofia, Bulgaria
tel. (+359 2) 987 72 90,
tel./fax (+359 2) 986 22 40,
office@stumejournals.com

ISSN 2367-749X
YEAR II, ISSUE 3 / 2016

EDITORIAL BOARD

CHIEF EDITOR
Prof. Dimitar Stavrev – Bulgaria

DEPUTY EDITOR:
Dr. Alexander Krugljakow, Germany
Prof. Sergey Dobatkin, Russia
Prof. Sergey Nikulin, Russia
Prof. Victor Anchev, Bulgaria

RESPONSIBLE SECRETARY:
Assoc. Prof. Tsanka Dikova, Bulgaria

MEMBERS:
Prof. Adel Mahmud, Iraq
Prof. Anna Proikova, Bulgaria
Prof. Bekir Sami Yilbas, Saudi Arabia
Prof. Dermot Brabazon, Ireland
Prof. Dipten Misra, India
Assoc. Prof. Eugeni Grigoriev, Russia
Prof. F. W. Travis, United Kingdom
Prof. Gennagiy Bagluk, Ukraine
Assoc. Prof. Ibrahim E. Saklakoglu, Turkey
Prof. Iis Sopyan, Malaysia
Prof. Ivan Parshorov, Bulgaria
Prof. Ivanja Markova, Bulgaria
Prof. Janez Grum, Slovenia
Prof. Jens Bergstrom, Sweden
Prof. Leszek Dobrzanski, Poland
Prof. Ludmila Kaputkina, Russia
Prof. Nikolai Dulgerov, Bulgaria
Prof. Omer Keles, Turkey
Prof. Plamen Danev, Bulgaria
Prof. Rui Vilar, Portugal
Prof. Rusko Shishkov, Bulgaria
Prof. Saleem Hashmi, Ireland
Dr. Sefika Kasman, Turkey
Prof. Seiji Katayama, Japan
Prof. Souren Mitra, India
Dr. Sumsun Naher, United Kingdom
Prof. Svetlana Gubenko, Ukraine
Prof. Sveto Cvetkovski, Macedonia
Prof. Ventsislav Toshkov, Bulgaria
Prof. Yovka Dragieva, Bulgaria
CONTENTS

MECHANICAL PROPERTIES OF DENTAL Co-Cr ALLOYS FABRICATED VIA CASTING AND SELECTIVE LASER MELTING

TECHNOLOGY LASER MODIFICATION IN MATERIAL SCIENCE NANOCOMPOSITES FOR METAL-POLYMERIC SYSTEMS
Sarokin V., Doc. cand. eng. Auchynnikau Y., Doc. cand. eng. Avdeychik S., Prof. dr. eng. Struk V. ......................................................... 8

MECHANOCHEMICAL TECHNOLOGY IN MATERIALS SCIENCE OF FUNCTIONAL METAL-POLYMER COMPOSITES

NEW MATERIALS FOR IMPLANTS OF THE HUMAN HIP JOINT AND TECHNOLOGY OF THEIR MACHINING WITH THE ACHIEVEMENT OF HIGH PRECISION AND QUALITY OF SPHERICAL SURFACES
Doctor of science, Prof., Turmanidze R., Undergraduate student Popkhadze G. .......................................................................................... 12

ON A PHYSICO-MATHEMATICAL MODEL FOR CONTROLLED FORMATION OF PERIODIC NANOSTRUCTURES AT SOLID SURFACES IRRADIATED BY FEMTOSECOND LASER PULSES
Prof. Dr.Sc.Math. Guseynov Sh.E., B.Sc.Phys. Žaimis U. ................................................................................................................... 17

AUTOWAVES OF LOCALIZED PLASTIC DEFORMATION ON THE YIELD PLATEAU AND ON THE WORK HARDENING STAGE

MATERIALS FOR ELECTROMAGNETIC INTERFERENCE SHIELDING
Ass.prof. PhD Kamelia Ruskova, Assos. Prof. PhD Ljudmila Taneva , Assos. Prof. PhD Alexandar Lirkov .................................................... 25

NEW CLASS OF COMPOSITE BRONZE, ARMED WITH STEEL DENDRITES FOR ANTIFRICTION TECHNIQUE
Prof. Dr. Potekhin B. A., Dr. Khristolyubov, A. S., Prof. Dr. Hernández Fereira A.A. .................................................................................. 28

INCREASING WEAR RESISTANCE OF MACHINES DETAILS FROM CARBON AND Cr – Mn – N STEELS OF THE COMPLEX BORIDE COATINGS
Prof. Dr. Chernega S., Poliakov I., Krasovskiy M. ................................................................................................................................. 31
MECHANICAL PROPERTIES OF DENTAL Co-Cr ALLOYS FABRICATED VIA CASTING AND SELECTIVE LASER MELTING

Assoc. Prof. N.A. Dolgov\textsuperscript{1,4}, Assoc. Prof. Ts. Dikova\textsuperscript{2,5}, Assist. Prof. Dzh. Dzhendov\textsuperscript{2,6}, D. Pavlova\textsuperscript{3,7}, Assist. Prof. M. Simov\textsuperscript{3,8}

\textsuperscript{1}Pisarenko Institute for Problems of Strength, Nat. Ac. Sci. of Ukraine, 2 Timiryazevskaya Str., 01014 Kiev, Ukraine
\textsuperscript{2}Faculty of Dental Medicine, Medical University of Varna, 55 Marin Drinov Str, 9000 Varna, Bulgaria
\textsuperscript{3}Medical College, Medical University of Varna, 55 Marin Drinov Str, 9000 Varna, Bulgaria

E-mail: \textsuperscript{4}dna@ipp.kiev.ua, \textsuperscript{5}tsanka_dikova@abv.bg, \textsuperscript{6}jendo_jendov@abv.bg, \textsuperscript{7}dianapavlova50@gmail.com, \textsuperscript{8}maksim_simov@abv.bg

Abstract: The aim of the present paper is to investigate the mechanical properties (hardness and tensile strength) of dental Co-Cr alloys fabricated via casting and selective laser melting (SLM). Two groups of metallic specimens (four-part dental bridges and standard tensile test specimens) made of Co-Cr dental alloys were produced by lost-wax casting and SLM processes. Vickers hardness distribution along the depth of the dental bridges as well as the Rockwell hardness and tensile strength of the samples were studied out. The hardness of Co–Cr dental alloys are dependent on the manufacturing technique employed. It was established that the average Vickers hardness of the samples, produced by SLM, was higher than that of the cast samples 382 HV and 335 HV respectively. The nearly even hardness distribution in the bridges, produced by SLM, and fluctuations of the hardness values along the depth of the cast bridges were observed. The Rockwell measurements confirmed the higher hardness of the SLM samples – 39 HRC in comparison with that of the cast ones – 33 HRC. The tensile strength is in good agreement with the hardness values. Due to the unique microstructure, the yield strength and tensile strength for the SLM samples were higher than those of the as-cast alloy.

Keywords: DENTAL Co-Cr ALLOYS, CASTING, SELECTIVE LASER MELTING, HARDNESS, TENSILE STRENGTH

1. Introduction

Cobalt–chromium based alloys have been widely used in various orthopedic implants as well as for manufacturing of metal framework of fixed dental prosthesis because of their excellent mechanical properties, high corrosion and wear resistance, and good biocompatibility. The chemical composition of dental Co-Cr alloys consists of 53–67% Co, 25–32% Cr, 2–6% Mo and small quantities of W, Si, Al and others [1]. Chromium, molybdenum and tungsten are added for strengthening of the solid solution. Due to the relatively large amount of Cr, dense passive layer of Cr2O3 with 1–4 nm thickness is formed on the surface of the details, determining the high corrosion resistance [2,3].

In proper alloying the microstructure of the dental alloys is composed mainly of γ-phase and carbides of the M23C6 type [3]. The high temperature γ-phase possess face centered cubic (FCC) lattice, while the room temperature ε-phase has hexagonal close packed (HCP) lattice [2-5]. The γ-phase determines ductility, while the ε-phase enhances the corrosion and wear resistance [7]. So, the properties of the Co-Cr dental alloys depend on the γ-ε ratio and the type, quantity and distribution of the carbide phase in the microstructure.

Most of the dental constructions are manufactured by lost-wax casting process which consists of many manual operations, leading to low accuracy and satisfactory quality. The new process of Selective Laser Melting (SLM) offers opportunity for overcoming the disadvantages of the casting process. In this technology layers of metal powder are fused into a 3D model by adopting a computer-directed laser [7-12]. The advantages of SLM over the traditional methods include production of personalized complex objects; manufacturing of parts with dense structure and predetermined surface roughness; controllable, easy and relatively quick process [7,13].

The SLM process characterizes with high heating and cooling rates of the melted layer as well as heating and solid state phase transformations in the underneath layers, which determine microstructure and properties quite different than that of the cast details.

Meacock et al. [14] reported that the microstructure of biomedical Co-Cr-Mo alloy, produced by laser powder microdeposition, is homogeneous comprised of fine cellular dendrites. The average hardness was 460 HV0.2, which is higher than the values obtained by the other fabrication process. Barucca et al. [15] investigated Co-Cr-Mo parts, produced by direct metal laser sintering. They established that microstructure consists of γ and ε phases. The ε phase is distributed as network of thin lamellae inside the γ-phase. The higher hardness is attributed to the presence of the ε-lamellae grown on the {111}, planes that restricts the dislocations movement in the γ-phase. Yanjin Lu et al. [16] investigated the microstructure, hardness, mechanical properties, electrochemical behaviour and metal release of Co-Cr-W alloy fabricated by SLM in two different scanning strategies – line and island. They established the coexistence of the γ- and ε-phases in the microstructure and nearly the same hardness – 570 HV for line-formed alloy and 564 HV for island-formed. Their research show that the results of tensile, hardness, density, electrochemical and metal release tests are independent of the scanning strategy and the yield strength of both samples meet the ISO 22764:2006 standard for dental restorations. Wen Shifeng et al. [17] investigated the influence of the samples orientation during SLM manufacturing process on the tensile strength. They established that the SLM specimens made along the vertical direction have higher tensile strength and elongation than those made along the horizontal direction, indicating significant anisotropic feature of SLM parts. According to them the molten pool boundaries have a significant impact and are the main reason for anisotropy and low ductility of SLM parts.

SLM is comparatively new process and the data about the microstructure and mechanical properties of the constructions, manufactured using t, are relatively scarce. The purpose of this study was to fabricate a Co–Cr alloy using the SLM process and casting process, as well as to investigate the microstructure, and mechanical properties.

2. Materials and methods

Materials and samples preparation

In order to understand the relationship between the process, microstructure, surface hardness and mechanical properties, tensile tests with specimens made of Co-Cr alloy were carried out. Two groups of samples – four-part dental bridges (Fig. 1) and tensile test specimens were prepared by lost-wax casting and SLM using Co–Cr dental alloys.

In order to obtain samples with sufficiently good repeatability at first a base model of 4-part dental bridge was made. It was used for manufacturing of silicone mold for production of wax models
and for generating of virtual 3D model. A silicone mold for manufacturing of wax models for the cast tensile test specimens was also fabricated, while the 3D model was created with SolidWorks software. The cast samples – bridges and tensile test specimens were produced by centrifugal casting of Co-Cr alloy “Biosil” with chemical composition, given by the producer (Table 1).

The SLM samples (Fig. 1) were fabricated directly from the virtual 3D models using SLM125 machine of the “SLM Solutions”, Germany. The machine is equipped with continuous Nd:YAG laser which worked with power 100 W and laser spot diameter 0.2 mm. The metal powder of Co-Cr alloy Co212-f ASTM F75 (Table 1) was melted in layers with 0.03 mm thickness unless the desired construction was obtained. During manufacturing process the laser at first scanned the outer contour of the layer of the first specimen’s part, next it hatched the area within the boundaries at an angle of 45° with a pitch of 0.13 mm. After that it passed to the same layer of the next specimen’s part, thus fabricating the whole layer. The laser path in manufacturing of tensile test specimen is shown on Fig. 2. The SLM technological regime, recommended from the company producer was used. The tensile test specimens have a thickness of 2.2 mm for cast alloy and 2.07 mm for SLM one, as well as width of 6 mm.

**Mechanical properties characterization**

The Vickers hardness (HV) was measured on preliminary prepared cross-sections of the bridges along depth of all their elements with 100 gf loading. The Rockwell hardness was measured in different areas of the polished tensile test specimens and the average value is taken. The microstructure of the cast and SLM bridges was investigated by optical microscopy.

Uniaxial tensile tests were performed using an FM-1000 testing machine at room temperature. The strain was measured using a strain gauge. Typical samples used throughout the study are shown in Fig. 3 and Fig. 4. The 0.2% offset yield strength (0.2%YS) and elastic modulus were obtained from the stress–strain curve. The fracture surfaces were evaluated after tensile tests using optical microscope.

### 3. Results obtained

**Hardness**

The Vickers hardness distribution along depth of the cast bridge is given on Fig. 5. It is uneven with high fluctuations of the values within the range of 222 HV0.1. The hardness distribution along depth of the SLM sample is much more even, as it is shown on Fig. 6 with lower average deviation of the values – 184 HV0.1. The average Vickers hardness of the cast samples is lower than that of the SLM ones: 335 HV0.1 and 382 HV0.1 respectively (Fig. 7-a). The average Vickers hardness values of dental bridges are in very good agreement with the average Rockwell hardness, measured on the tensile test specimens. The Rockwell hardness (Fig. 7-b) of the cast sample is 33 HRC (331 HV), which is lower than that of the SLM samples – 39 HRC (382 HV). The Rockwell hardness measurements confirm the lower hardness deviation of the SLM samples.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Chemical composition, mass %</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Co</td>
</tr>
<tr>
<td>ASTM F75</td>
<td>Bal.</td>
</tr>
<tr>
<td>Biosil, Degudent</td>
<td>64.8</td>
</tr>
<tr>
<td>SLM Co212-f</td>
<td>65.2</td>
</tr>
<tr>
<td>ASTM F75</td>
<td></td>
</tr>
</tbody>
</table>

**Table 1. Chemical composition of the alloys used.**

### Table 2. Mean of mechanical properties of cast and SLM alloys.

<table>
<thead>
<tr>
<th>Variable</th>
<th>BioSil F</th>
<th>SLM</th>
</tr>
</thead>
<tbody>
<tr>
<td>Yield strength (MPa)</td>
<td>410</td>
<td>720</td>
</tr>
<tr>
<td>Elastic modulus (GPa)</td>
<td>209</td>
<td>213</td>
</tr>
<tr>
<td>Hardness HRC</td>
<td>33</td>
<td>39</td>
</tr>
<tr>
<td>Hardness HV 10</td>
<td>335</td>
<td>382</td>
</tr>
</tbody>
</table>

**Tensile test**

Figure 8 shows the typical nominal stress–strain curves of the SLM specimen and as-cast alloy. On the basis of such curves, the mechanical behavior can be recorded, including elastic deformation, and plastic deformation with the yielding process. Table 2 is a summary of their mechanical properties. The SLM specimens showed higher 0.2% yield strength compared to the as-cast alloy – 720 MPa and 410 MPa respectively. The yield strength of SLM specimens meets the standard ISO 22674:2006 for dental restorations (> 500 MPa).
Fig. 5. Hardness distribution along depth of the cast sample.

Fig. 6. Hardness distribution along depth of the SLM sample.
4. Discussion

The properties of the Co-Cr-Mo alloys depend on the microstructure, its morphology and composition, γ-ε ratio, presence of carbides and intermetallic precipitations. Mechanical properties are affected by many factors, such as phase, grain orientation, grain boundary conditions, defects, etc.

The microstructure of the cast sample (Fig. 5) characterizes with dendrite morphology and carbides with round shape and small sizes, situated inside the grains. Our previous investigations [18] showed that the dendrites are composed of γ phase with lower strength, while the interdendritic regions consist of γ-phase/intermetallic eutectic with carbides, defining the higher hardness. The prevailing volume fraction of the lower strength γ phase is the main reason for the lower hardness of the cast samples, while the microstructural inhomogeneity is responsible for the higher fluctuations of the hardness values (Fig. 5).

The SLM process characterizes with high heating and cooling rates, leading to fine grained microstructure of the solidified layer. As the heat is lead away through the solid body, phase transformations run in the underneath layers heated above the transition temperatures, resulting in more homogeneous microstructure. Increasing the samples volume during the manufacturing process reduces the cooling rates which lead to slightly hardness decrease (Fig. 6).

The microstructure of the SLM sample (Fig. 6) possesses high corrosion resistance and could not be etched by immersion in any reagent, given in the references. Pores with different sizes, elongated along the direction of the layers melting, as well as unmelted powder can be observed in the whole volume. Our previous investigations [18] showed even distribution of the chemical elements in the dense areas of the SLM sample, proving the more homogeneous microstructure. In samples, produced by SLM, the higher hardness is attributed to the homogeneous microstructure with fine morphology [14] and the higher volume fraction of ε phase due to the incomplete γ-ε transformation [15], which is defined by the process peculiarities.

Mechanical properties of the SLM specimens and casting ones were investigated and compared. The SLM alloy possessed improved mechanical properties in comparison with that of the cast alloy. Yield strength and tensile strength of the SLM specimen was improved due to the unique microstructure. The higher tensile strength of the SLM specimens may be due to the finer grain size, cellular dendrite, and elongated precipitates. Also residual stress in the SLM specimen generates the residual strain [12] and affects on the tensile strength. Additionally, the presence of residual stresses during sintering is possible reason for the increased hardness in the SLM.

5. Conclusion

This study is focused on the microstructural and mechanical characterization of Co-Cr alloys fabricated using casting and selective laser melting. Within the limitations of this study, the following conclusions can be derived:

- Co–Cr dental alloys fabricated via casting or SLM techniques show significant differences in strength and hardness.
- The average Vickers hardness of the samples, produced by SLM, was higher than that of the cast samples – 382 HV and 335 HV respectively. The Rockwell measurements confirmed the higher hardness of the SLM samples – 39 HRC in comparison with that of the cast ones – 33 HRC.
- The SLM specimens showed higher yield strength compared to the as-cast alloy – 720 MPa and 410 MPa respectively.
Compared with the currently used cast alloy, the SLM alloy possessed improved mechanical properties. Co-Cr dental alloy fabricated with SLM is a promising alternative to conventional cast alloy for metal ceramic restorations.

6. References
Abstract: Are investigated mechanisms and kinetics of transformation polymeric substrates and of disperse particles fillers and modifiers exposed defocused laser radiation with energy in the range of 1.5 to 6 J. Installed effect of hardening the thermoplastic matrices (polylefins, polyamides, of polyesters) when exposed to short laser pulses on a film sample thicknesses of 50-200 microns. The methods of structural analysis (IR spectroscopy, X-ray diffraction, DTA) showed, that the predominant mechanism of manifestation hardening effect is the formation of nano-sized structures in the film sample volume due to leakage of recrystallization processes. When exposed to laser radiation to disperse and fibrous particles of silicates, graphite, shungite, carbon fibers is implementing a complex effect of increasing the dispersion and the formation of advanced morphology of the surface layer with an increase in the proportion of nano-sized fragments whisker and lamellar habitus. When laser irradiation of fibers fragments formed developed morphology of the surface layer, due to the occurrence of thermal degradation processes that lead to the appearance of cracks and of globular indentations. Spectroscopy method of thermally stimulated currents (TSC - spectroscopy) established the effect of changing the energy state of the surface layers of substrates and particulate modifiers, which are characterized by extreme values of magnitude TSC - currents in the temperature range 193-523 K. The combined effect of energy and morphological factors provides the effect of increasing the activity of modifying components subjected to the laser processing. Changing the energy state of the surface layer of the polymer substrate has a beneficial bactericidal effect, increases the effectiveness of antiseptic treatment of medical devices used in medical practice. Are presented examples of the practical use of the established laws when creating polymer composites for metal-polymer systems for various applications.

KEYWORDS: MODIFICATION, LASER INFLUENCE, MORPHOLOGY, SURFACE LAYER, ENERGY STATE, STRUCTURING MATRICES

1. Introduction

The development of effective compositions of composite materials with desired performance parameters is one of perspective directions of creation of static (adhesive) and dynamic (tribological) Metal-Polymer Systems [1, 2].

Clearly a statement that the determining criteria for the selection of the functional material components are the degree of their influence on the mechanisms and kinetics of interfacial physico-chemical and tribological processes at the interface of "matrix-filler" and "composite-substrate", which is determined by the activity, assessed by parameters energy characteristics [2].

Among the most promising parameter control technologies and energy morphological characteristics of the metal components and metal-polymer systems include laser radiation with a certain energy and time parameters. Laser irradiation allows not only to carry out the dispersion of semi-finished products as a result of ablation and destruction, but to change the parameters of the structure and energy characteristics of the dispersed particles and substrates directionally adjusting mechanisms of contact interactions [3, 4].

The aim of this work was to study the morphology and energy characteristics of the components Metal-Polymer systems exposed to laser radiation.

1. Materials and methods of research

For the research were chosen components, which widespread in the technology of functional materials based on polymeric matrices - polylefins (HDPE, LDPE), polyester (PET, PBT), polyamides (PA6, PA6.6), fluoropolymers (PTFE). As used modifiers disperse silicon particles (clay), carbon containing (UFDG, graphite) and combined (shungite) compounds. Modifying components was performed in air or in a liquid medium (aqueous solutions) using a short-pulse laser radiation with energy (0.5 ÷ 0.75) J. and wave length λ₀ = 0.532 microns. Modified components used to form the adhesive system and coating producing composite materials.

2. Results and discussion

System analysis of studies on various aspects of the technology of laser exposure on the components of Metal-Polymer Systems of different types [3, 4], indicate multifactorial influence of concentrated streams on the structure and energy parameters of the resulting products. At the same time, irrespective of the flow rate and the method of delivery (concentrated or defocused beams continuous or discrete action), there are structural changes at different levels of organization and energy parameters of the surface layer, which is caused by the occurrence of recrystallization processes, the formation of defects, energy absorption, degradation, ablation in the surface layers of the modified object. Model studies of the morphology of the surface layer films of thermoplastic polymers show a significant relief change with the formation of nano-sized components (Figure 1).

![Fig. 1 The surface morphology of film samples LDPE (a, d), PP (b, e), PET (c, f) the source (a, b, c), treated with pulsed laser light with a 2.0 W/cm² power density (d, e, f)](image-url)
Formation of nanoscale components in the surface layer characteristic of all types of thermoplastic samples subjected to treatment with pulsed power density (2.0-4.5) W / cm².

If the laser beam is not only changing the morphology of the surface layer, but its energy state (Figure 2), as measured by the magnitude of thermally stimulated currents. The probable mechanism of manifestation of the electret state is the formation of radical products of thermal degradation of macromolecules under the action of radiation, the amount of which depends on the amount of energy absorbed by the sample.

Of particular interest are the processes of functional modifying component materials in liquids of different composition. With such a technology to implement exposure manages modifying surfaces juvenile products laser dispersing functional components introduced into the liquid-phase medium, as electrolytes which can serve for the galvanic coating, solutions of surfactants, solutions of salts of polyvalent metals and etc.

Exposure to the laser pulse on a substrate of different composition causes the destruction of the surface layer by the mechanism of thermal degradation and ablation (Figure 3).

Have developed the morphology of the surface layer in the ablation of particles ensures the implementation nanostate, which has a significant impact on the mechanisms and kinetics of interfacial interactions in composite materials and metal-polymer systems.

For example, the products of ablation of PTFE and graphite electrode introduced into the electrolyte for the production of coatings based on chromium, can not only change the rate of formation of the coating composition, but also improve its parameters tribological and corrosion resistance.

The characteristic relief of nanocomposite coatings based on modified products ablation chromium (Figure 4) contributes to the formation of sustainable boundary lubrication in heavily loaded friction units. The presence in the structure of the coating of nanoparticles of graphite and PTFE significantly reduces the coefficient of friction during its operation without external supply of lubricant.

Of particular interest as a modifier are fluorine-containing oligomers (FCO), which form the structure of the surface layer which is resistant to corrosion and mechanical wear.

3. Conclusions

Systemic investigations of the mechanisms of formation and morphology of the energy state of the components of different composition and structure when exposed to laser light indicate the presence of the general laws, leading to the formation of nanocomponents, which ensure the implementation of the phenomenon nanostate.

The components activated by laser radiation of different energies are effective for modifying polymer and metal matrix composite materials to create a set of performance parameters. A promising area is the use of laser modification optically transparent liquid-phase fluids containing functional modifiers are capable of adsorbing on the surfaces of juvenile dispersion and ablation products.

4. References

Abstract: There were studied mechanisms interfacial interactions in composite systems based on dispersed particles of polymer materials (polyolefins, polyamides, fluorcontaining polymers) and modifiers of various composition and structure in the combination process in the units with intense mechanical action. There were established the effects of formation of products of mechanochemical interaction of active centers of modifiers particles and radical fragments of macromolecules degradation. The obtained results allow changing the technological paradigm of the creation of high-strength and highly filled engineering nanocomposites based on high-molecular matrix.

KEYWORDS: DISPERSED PARTICLE, POLYMER MATRIX, MECHANOCHEMICAL INTERACTION, NANOCOMPOSITE.

1. Introduction

In a wide range of engineering materials a special place occupy composites of various purposes, which are obtained by modifying the matrix polymer, oligomeric and combined bindings by target components based on silicon, carbon, metal-containing compounds with a given dispersion and form [1-4]. In all cases the use of modifiers of high-molecular matrix generalizing (basic) criteria of their reasonable choice is given effectiveness of action necessitates technically significant effect of improving the parameters of deformation strength, tribological, thermal, adhesive, etc. service characteristics with minimal economic costs, taking into account in material, energy capacity and staffing processes of production and consumption of the target.

An obvious is the condition of fulfillment of this criterion, suggesting the achievement of optimal structural state at a given level of organization of the composite material at a particular combination of matrix binder, modifier type and technology of its combining and processing.

The purpose of this study was to assess the prospects of using mechanochemical technologies to produce nanocomposite materials based on polymeric materials.

2. Methods of research

For research have been used thermoplastic matrices dispersed particles and of silicon containing modifiers. The combination of the components was carried out in blending machine AGO-2 type under acceleration 2-20 g.

3. Results and discussion

An analysis of the literature sources devoted to materials science and technology of polymeric composites [1-4], shows promising use of modifiers that are in the nanometer range, so-called nanosized particles of different composition, structure and technology. In the presence of a large number of studies on the mechanisms of modifying action of nanoparticles in the polymeric, oligomeric and combined matrices, it is necessary to underline the ambiguity of the results and the lack of common concepts that define the dimension and concentration ranges of the optimal effective operation of the dispersed particles in the matrix binders, differing in molecular weight, chemical structure of macromolecules, peculiarities of structural organization under the influence of technological factors - temperature, pressure, exposure time, etc. The most effective multipurpose modifiers are natural silicon compounds - clays and zeolites, fine products of detonation, thermo gas dynamic and plasma chemical synthesis – nanoceramics, sialons, nano-diamonds, diamond-containing burden, fullerenes, nanotubes, thermally split graphite, mica, clays, and nanosized particles of metals and oxides obtained by thermolysis of metal-containing precursors in molten thermoplastics [1-4].

A particular interest presents the particles of inorganic natural substances, which, when exposed to a special technology to forms nanosized assemblies of single particles or components with a thickness of single plates around 1 nm - clay, mica, graphite etc. (figure 1). At the introduction of such fillers in an amount of 3 – 5 wt. % in a polymeric matrix such as aliphatic polyamides provided 1.5-2 times increase of strength characteristics of the composite material. The mechanism of the observed effect is associated with the formation of graft polymers on the active surfaces of nanoplates of layered minerals [4].

Among the especially promising areas of target modification components is their mechanochemical activation for given conditions of energy impact [1]. Due to mechanical stresses of certain intensity is ensured not only the formation of the active centers of the surface layers of the matrix and the modifying components but their interaction with the formation of the combined product of the boundary layer the optimal structure (figures 2-3).
Using the method of scanning electron microscopy studied the morphology of the particles of mechanically activated polyamide 6 and kaolinite before and after exposure to ionizing radiation (Figure 3, a-b).

The initial particles (polyamide, kaolinite) until the mechanical activation operation had a dispersion, which is in the range of 50 microns to 200 microns. The morphology of the particles in the delivery condition has a shape characteristic of whisker particles. A joint mechanical activation of particles of polyamide 6 and kaolinite leads to changes in the morphology and the formation of particles with a morphology close to a globular form. This conclusion is confirmed by the calculations to determine the form factor values for the particles investigated.

At mechanical activation of kaolinite mixed with polyamide is going the process of disordering kaolinite structure due to rupture of (Al)-OH bonds and respectively bonds violations in octahedral and tetrahedral grids of layer. Meanwhile are formed the surface active centers. Presumably the active centers of a basic character of silicate interact with the active centers of PA6. During the activation process produced a polymer composition in which the polymer and kaolinite chemically linked. Water formation during mechanical activation confirms the absorbance increase in 3550-3350 cm\(^{-1}\), whereby the \(\nu(\text{-NH-})\) strip of amide group becomes asymmetric.

4. Conclusion

The resulting products of mechanochemical combining were used as the polyamide matrix modifier (PA6) and as an intermediate in the obtaining of dispersible powders for coating metal substrates. Studies have shown that the mechanochemical combination of components achieves a synergistic effect increase of parameters of deformation strength and adhesion characteristics, with a marked (1.5-2.0 times) increase in the wear resistance of coatings during operation without an external supply of lubricant. Studies indicate the prospects of the use of mechanochemical technologies in functional materials science.

5. Literature

NEW MATERIALS FOR IMPLANTS OF THE HUMAN HIP JOINT AND TECHNOLOGY OF THEIR MACHINING WITH THE ACHIEVEMENT OF HIGH PRECISION AND QUALITY OF SPHERICAL SURFACES

Doctor of science, Prof., Turmanidze R.¹, Undergraduate student Popkhadze G.²
Faculty of Transportation and Mechanical Engineering ¹, ² – Georgian Technical University, Tbilisi, Georgia, inform@gtu.ge

Abstract: In view of the fact that the endo-prosthesis heads of human hip-joint are operated in extreme conditions, in respect of load, the selection of corresponding material and also increase of precision and quality of machining of spherical surfaces is rather topical task.

In the submitted work are reviewed the problems connected with definition of the influence degree of orientation of the sapphire crystal on its workability during diamond grinding with a butt of the ring and elaboration of the perspective, original scheme of formation of the incomplete spherical surface, particularly, of the sapphire head of endo-prosthesis of the human hip-joint.

KEYWORDS: SINGLE CRYSTAL SAPPHIRE, ANISOTROPY, GRINDIBILITY, ENDOPROSTHESIS, PRECISION GRINDING FORMING, SPHERICAL SURFACE.

1. Introduction

Endo-prosthetics is effective and often the only method of the function restoration of the human joint. It is established that every year in the world are made about a million operations of exchange of the human hip-joint.

The endo-prosthesis heads of the human hip-joint from the point of view of the character and value of load are operated in extreme conditions. The contemporary joint endo-prostheses consist of acetabular (cup) and hip component (leg) and also the head from metals or ceramic materials on their base. At present there is an acute problem of creation of wear resistant inert materials for implantology.

Therefore, in each specific case the selection of the necessary material with corresponding physical – mechanical characteristics and also increasing of precision and quality of the most significant part of endo-prosthesis – spherical surfaces is rather actual task the acuteness of which intensively grows in recent years. The number of used endo-prostheses is some tens of millions of pieces a year and the statistics shows that unfortunately this number increases every year.

The production of implants from bio-ceramic materials is rather profitable direction. Up to now a powerful industry of fabrication of implants, tools and accompanying materials has been created and the western market of this product is evaluated in 2,5-3 billion Dollars a year. The basic developing and producing countries of implants are: USA, Japan, Germany, France, Great Britain, Russia, Italy, South Korea, countries of Asia region and other countries.

It is conditioned by the fact that if earlier the necessity of similar operations was caused by the age factor of the man or traumatology fractures in recent two decades abruptly increased the number of patients at young age of 30-40 years of both men and women without any injuries and fractures. In opinions of physicians the principal reasons of it are non-active way of life of youth, composition of contemporary artificial food products and metabolic disorder. All the above-mentioned reasons determine the number of used endo-prostheses in tens of millions of pieces a year.

The medical practice proves that the repeated prosthetics of the human hip joint is connected with big problems. In many cases the implementation of such operations becomes practically impossible. Therefore, the durability of the endo-prosthesis of the human hip joint to the end of the patient’s life especially at young age has especially significant meaning.

The contemporary elaborations of the endo-prosthesis structure are directed to exchange of chirulen by ceramic material that leads to the change of the endo-prosthesis structure, development of new ceramic materials with the improved physical and mechanical characteristics, development of processing of internal and external spherical surfaces, development of diamond tools providing the high quality of the machined surface, definition of the optimal wear resistance of the pair of materials for manufacture of the endo-prosthesis etc.

For nowadays in the world practice these heads are manufactured from various alloys, composite materials and ceramics which mainly are isotropic materials. Therefore, the data of the above-mentioned works do not give the necessary information on machining of anisotropic materials, particularly artificial crystal of sapphire (Fig. 1).

Together with indisputable positive qualities the contemporary structures of endo-prostheses from metal, ceramics polymers have certain defects. Among them there is an insufficient biological inertness and redundant wear of components of the friction pair that lead to short duration of work of the artificial hip-joint. Besides that as the result of wear of the material in the friction pair of endo-prosthesis hinge toxic and onco-dangerous products of the dissociation in various organs and tissues accumulate that in 30-40% leads to hard complications and requires complex disabling interferences.

The circle of materials that satisfy the criterion of biological compatibility is rather limited. Because of it and requirements of the resource of articles in condition of action of alternating loads, corrosion active environment titanium and its alloys as the material for fabrication of endo-prostheses have the advantage over Co-Cr-Mo alloys.

Fig. 1. Hip heads from sapphire (a), Zirconium ceramics (a), b - Stainless steel, c - Biologically pure titanium alloy.
The wear of heads from zirconium ceramics is 7-9 nm a year while titanium 105 nm a year. Coming out of this the fabrication of implants from bio-ceramic materials is a perspective direction of development of the science intensive technologies of creation of materials and their machining. Most bio-compatible with the human organism, wear resistant and durable material for fabrication of the above-mentioned article is the artificial mono-crystal of sapphire.

For nowadays basically the circle of materials for fabrication of pairs of bearing surfaces of endo-prostheses of joints with minimally possible number of products of wear has been determined. Such pairs of bearing surfaces with excellent characteristics of wear resistance are the friction pairs of ceramics-ceramics, metal-metal and polyethylene with a high degree of cross-section links in combination with ceramics or metal. Together with that the basic unresolved problem of the last decade was the development of bearing surfaces that could endure much higher loads with young and movable patients. The surfaces that are being investigated at present in laboratory conditions due to their hopeful characteristics of wear are ceramic matrix (82% of aluminum oxide, 17% of zirconium dioxide, 0.3% chrome oxide), zirconium dioxide and ceramics in pair with cobalt-chrome alloy.

Together with that sapphire representing the mono-crystal of aluminum oxide as a material for bearing surfaces possesses unique inertness including electrolyte passiveness, perhaps the best of the well-known materials, bio-compatibility, corrosion resistance and hardness. Resistance of sapphire towards any acids and alkali is rather higher than that of metals and even poly-crystalline oxide of aluminum. Probably because of it sapphire doesn’t change the immune status of the patient. If metals and poly-crystalline materials used for the bearing surfaces have different speed of wear of the micro-sites it leads to the coefficient increase of the pair friction and wear increase then this effect with sapphire is absent.

Tribological investigations of the friction pairs of materials were conducted by the scheme of rotational friction with the contact geometry of “ball on disk” type. The flat disk was manufactured from the investigated material and the ball from the material of counter body (Fig. 2).

![Fig. 2. Schematic arrangement of rotational friction – a, and example of counter bodies – b.](image)

The obtained dependencies of the friction force of pairs: sapphire-sapphire, sapphire-ruby from the time of tests (number of cycles) has mainly non-monotonous abrupt character. From them one can mark out three stages of wear, namely: running-in wear – creation of the work roughness and necessary supporting surface on frictional surfaces. At the moment of beginning of work the contact of bodies takes place in the point, accordingly the specific load is sufficiently high that leads to the abrupt growth of the friction force and as a consequence of wear of the surface material.

Normal wear is after achievement of the surface wear meaning, at which the optimal meanings of characteristics of supporting surface are reached, the stable process of friction of bodies has place with gradually decrease force of friction.

Catastrophic wear is in the process of friction of counter bodies of the wear products which fill the pockets on the friction surface and because of the weak (drop) feeding of the Ringer solution into the area of friction the wear products accumulate occupying all the vacant place in the pockets and as a result of this take part in the process of wear of the investigated surfaces as a free abrasive.

The pointed out above stages of wear can have different length of time or generally to be absent in the wear process.

From the dependencies follows that the best results in relation of the friction force to the pressing force shows the friction pair of sapphire-titanium with orientation of the crystalline lattice on flatness 0001, linear wear of the friction pair of sapphire-titanium with the orientation of the crystalline lattice on the flatness 0001 and sapphire-ruby.

Titanium and alloys on its basis are widely used in medicine as implants and other articles. From the point of view of bio-compatibility for the implants working for a long time in the live organism the use of titanium is preferable that unlike alloys does not contain alloying additives harmful for the live organism. However, titanium in its usual state has low mechanical properties in comparison with its alloys. This problem was solved by means of formation of nano - and composition structure in technically pure titanium.

The use of the friction pairs of sapphire/titanium consisting of non-toxic materials would help us to solve the pointed out problem. However, for nowadays it is considered that from titanium and its alloys is impossible to manufacture a friction pair because of their high inclination to a contact snatch and as a consequence increased wear during friction. This property makes dangerous the use of titanium in the friction pairs (Fig. 3).

![Fig. 3 Experimental specimens of titanium head and sapphire cup for endo-prosthesis of the hip-joint.](image)

Thus, the work objective is the creation of a new bearing connection of endo-prosthesis of the hip-joint having the improved quality at the expense of use of hardened sapphire and biologically inert technically pure titanium with modification of the surface layer of IPD and subsequent
nitrating as materials improving tribological properties of the connection.

For conduct of investigations of machinability of the sapphire crystal the method of Low Temperature Precision Grinding (LPG) has been chosen developed at the Department of Mechanical Engineering of the Georgian Technical University as a version of progressive methods of diamond grinding of hard and brittle non-metallic materials.

In Fig.4a the machined parts set on the cassette in separators or by other method of mounting, for example, gluing carry out rotational movement with angular speed of \( \omega_2 \) and the grinding ring with speed of \( \omega_1 \) in the same direction. In the cutting area the pressing is carried out by force \( P \) of the machined surfaces of parts to the work surface of the grinding ring.

The experimental investigations were conducted on the laboratory setting equipped with the special precision head (Fig. 4b).

By the data analysis of complex, all-round experimental investigations conducted by us one can make the following conclusion:

By other equal conditions of machining the most intractable is orientation 0001. For all the tested diamond rings the ratio of meanings of the linear removal of material \( q \) is within \( q_{0001}/q_{1010}=0,25...0,5 \), at that, \( q_{1012}/q_{1010}=0,75...1 \).

The productivity of machining grows in the range of cutting speed of \( V=1...6 \) m/sec while at the further increase of the cutting speed to 12m/sec it remain constant.

With increase of the pressing force \( P \) in all the investigated range the productivity of machining grows, however, in the interval of \( P=1000...1500 \) kPa the productivity growth considerably decreases.

From characteristics of the diamond tool on productivity by the prevailing way influence granularity and bunch of the diamond tool. The concentration influence is insignificant. With the increase of the grain size within \( d=14/10...28/20 \) the productivity grows 1.5...2.5 times. The maximum productivity of machining is provided by the tool on ceramic bunch (\{0001\} – 130mkm/min; \{1010\} – 300mkm/min; \{1012\} –250mkm/min), then on metallic (\{0001\} – 50 mkm/min; \{1010\} – 200mkm/min; \{1012\} – 170mkm/min) and organic (\{0001\} – 30mkm/min; \{1010\} – 120mkm/min; \{1012\} – 110mkm/min). The tool on the ceramic bunch works in the mode of self-grinding.

From characteristics of the diamond tool the granularity and material of the tool bunch influence in prevailing way on the surface quality. With the grain increase in the investigated range the height of unevenness \( R_z \) grows within 1…1.5 classes and the depth of the violated layer \( H \) 1.5...2 times. In other equal conditions of machining on orientation (0001) higher quality of the surface is reached than on the rest two. The difference is in 1…1.5 classes of roughness. By this indication the best results are shown by the diamond rings on organic bunch. So, for example, on the diamond rings on bunches BC-11 and organic special are obtained the following results: \( R_z=0,25 \) mkm; \( t_{03}=35...45\% \); \( H=2...5 \) mkm. At that the meaning of parameter \( R_z \) is in order below than parameter \( t_{03} \) – 1,5 times higher of parameter \( H \) – 3…5 fewer than the meanings of corresponding parameters that are obtained on the diamond rings on ceramic and metallic bunches.

The influence character of the process factors on output parameters for the chosen orientations of the sapphire crystal ((0001), (1010), (1012)) is constant.

By study of morphology of the machined surface the possibility of cutting the sapphire material by plastic deformation of the removed layer at low cutting speeds \( V=1...3 \) m/sec with the least depth of the violated sub-relief layer has been proved (Fig. 5).

It is known that machining of vitreous materials, particularly, the sapphire crystal by means of plastic deformation of the removed layer instead of fragile destruction – dispersion, the pledge of acquisition of the machined surface practically without hereditary defect – without sub-relief layer. Value \( H \) appeared to be the least namely on these specimens of sapphire. The obtained meaningful result requires individual investigations the conduct of which is being planned.
The development of a new or updating of the existing technological process of machining of the sapphire head sets the actual task of creation of new highly effective schemes of formation. The optimization criteria of technological operations such as productivity, indices of the surface quality and precision of machining determine the place of new schemes-methods of formation in technological process taking into account their advantages.

The schematic arrangement of the spherical head of endo-prosthesis is shown in Fig. 6. The partial sphere is determined by its radius and angle $\beta$ of a spherical segment. There are several ways of machining of the spherical head.

The closest to the LPG process in kinematics is the scheme of grinding of the partial sphere with use of the end grinding ring. The machined part rotates with an angular speed of $\omega_2$ and $\omega_3$ around axes 2 and 3 subsequently (Fig. 7). This kinematics forming the incomplete surface of the sphere is taken as the basis of development and possible realization of more effective schemes of grinding of spherical heads of endo-prosthesis taking into account the kinematics and other positive features of the LPG method [1–6].

In Fig. 8a are shown eight subsequent positions of the spherical head in one cycle. For clarification all the positions are equally distributed along the ring although such subsequent positions are in the same place M. This scheme enables also carry machining in “machining stations” around the ring.

The main defect of this scheme is a complexity of the grinding ring and difference of speeds in points A and B. The latter can be compensated in the same way as in the LPG process. Comparatively simple schemes of formation of the spherical heads of endo-prosthesis by using commercially available shapes of the ring are shown below in Fig. 9 a, b, c and d. The schemes presented in Fig. 9 a, b and c differs only by the shape of used rings. The kinematics of formation of the partial spherical head is the same as the first scheme presented in Fig. 7 and 8.

As for the case described in Fig. 8 the partial spherical head rests between two grinding rings but preferable are three (Fig. 9d) that accordingly consist of two or three terminally located elements. For effective use of full space of abrasives on the used rings is suggested the implementation of coordinated in time harmonic movement of rings and machined sphere within the required distances as it is shown in Fig. 9 with arrows.

The diameters of grinding rings are unlimited when the grinding mode is carried out with two rings. In case of grinding with three heads the sphere radius ($r$) and maximum radius of the ring (R) are interdependent. The mode of three rings provides a high stability of the grinding process. In subsequent studies will be seen more detailed development of structural schemes of
formation of spherical heads and carried out experiments with the real spherical heads for endo-prosthesis.

2. Conclusion

1. Scale of influence of single-crystal sapphire of the crystallographic flat orientation and the grinding conditions in the mode of removal of material, surface finishing and the state of sub-surface layer studied during low-temperature precise grinding have revealed that workability of single-crystal sapphire considerably depends on the crystallographic flat orientation. The relative values of removal of the material against the specimen with crystallographic flat orientation (1010) are within $q_{1010}/q_{010}=0.25...0.5$, and $q_{1012}/q_{010}=0.75...1$. In other equal conditions of the LPG process the high quality of the ground surface is achieved for crystal orientation (1010). The difference with two other orientations of the crystallographic flatness is within 1...1.5 class of roughness.

2. By the studies of morphology of the ground surface was proved the grinding possibility by single-crystal sapphire in compliant mode, i.e. removal of the deformed layer by cutting without cracks at the low speed of cutting. In such conditions of machining was detected the least depth of the faulty sub-surface layer.

3. The structural schemes of formation of the partial spherical heads of endo-prosthesis for human hip-joint need a further analysis and optimization with the aim of designing of advanced technological processes and the prototype of grinding machines.

3. Literature


Fig. 9. Possible versions of grinding of the endo-prosthesis during the use of commercially available usual rings.
ON A PHYSICO-MATHEMATICAL MODEL FOR CONTROLLED FORMATION OF PERIODIC NANOSTRUCTURES AT SOLID SURFACES IRRADIATED BY FEMTOSECOND LASER PULSES

Prof. Dr.Sc.Math. Guseynov Sh.E.1,2, B.Sc.Phys. Žaimis U.1
Faculty of Science and Engineering, University of Liepaja, Liepaja, Latvia 1
"Entelgine" Research & Advisory Co., Ltd., Riga, Latvia 2
sh.e.guseinov@inbox.lv, uldiss.z@inbox.lv

Abstract: In this paper, we propose a physico-mathematical 3D model to study some of the basic features of the interaction of radiation with solid materials (metals, semiconductors, insulators) under the action of femtosecond laser pulses. The constructed model is a parametric model that, in particular, takes into account the dependence of the physical and chemical characteristics of the periodic nanosurface structures controlled laser irradiation parameters the polarization, angle of incidence, the energy density, wavelength, etc. The developed model is meant for the excitation of periodic nano-surface structures by electric waves which are in the process of irradiation and periodically amplified and attenuated, i.e. it holds periodic interference.

Keywords: PERIODIC NANOSTRUCTURE, LASER PULSE, MATHEMATICAL MODEL, STABLE SOLUTION

1. Introduction

The synthesis of periodic nanostructures on various solid surfaces together with femtosecond laser impulses is one of the most perspective methods in the field of laser ablation on the metal surfaces, metal-like materials and dielectrics. In particular, the use of femtosecond lasers has several advantages. First of all, the effect of these impulses on the target leads to ultra-fast heating, which then reaches a very high quality of material processing. Second of all, the next two very crucial aspects provide unique conditions for the ablation of material together with the femtosecond laser impulses (see, for instance, [1]-[3]). Now it is vital to observe that the amount of time it takes after the laser impulse from the heated zone is negligible so that the laser energy is absorbed in the targeted and localized area that can be controlled.

Currently the laser synthesis of periodic nano-surface structures is successfully applied in increasing the ultra-thin materials that are often used in the microelectronics and aerospace materials, for colour coding, for local chemical modification of surfaces, to excite Raman light scattering, and to improve the strength of ultra-thin materials (see, for instance, [4], [5] and the corresponding references therein). Nevertheless, there are no adequate complex physical and mathematical models that are based on rigorous theoretical calculations; no strictly mathematically based concept can make accurate predictions in the evolution of both the geometry and characteristics of the synthesized periodic nano-surface structures on surfaces of various metals and metal-like materials under the influence of femtosecond laser impulses with different physical and technological conditions.

For an accurate description of the mechanism of synthesis of periodic nanostructures on the surface of semiconductors by the femtosecond laser impulses, there appears different approaches that mostly use the same method of studying the analysis of various surface instabilities. Notice that such analysis is not able to fully identify the particular nano-surface periodic structures (see, for instance, [6], [7]), as the absorption of the femtosecond pulse geometry of the surface of the material hardly changes. Thus, this is considered as a problem of the non-uniform distribution of the laser energy on the treated nano-surfaces; optical properties are instantly changed in the course of photoexcitation and the melting of the electronic subsystem.

Simultaneously, the developed models for solving this problem apply only to a perfectly flat or rough surface with very low relief modulation amplitude ([9], [8]). In particular, when the ratio between the relief modulation and the laser wavelength is negligible, and in both cases, the following restrictions apply: the number of laser impulses must not exceed 10 and the periodic values of the nano-surface structures should be in the range 1-1.1 (see, for instance, [9], [10]).

The authors of this work have built from scratch 1D-3D mathematical models for the study of some of the main features of the interaction of radiation with solid materials (taking as an example copper) under the action of femtosecond laser pulses. The models follow the parametric model, which incorporates the dependence of physical and chemical characteristics of the periodic nano-surface structures (periodic nanoscale particles) from the controlled parameters of laser ablation: from polarization, the angle of incidence, the energy density, of the wave length, etc. All the new models are expressed in the form of linear and non-linear non-stationary partial differential equations with linear and nonlinear boundary conditions of mixed type (the Neumann-type and Newton-Robin type).

Furthermore, the authors of this paper developed an analytical and numerical algorithm for finding stable solutions in 1D approximation for constructed 3D model. However, due to restrictions on the presented volume of this work, we will skip the output constructed 1D-3D physical and mathematical models and give only the setting of the "ready-made" 3D model. Therefore, we will present the developed analytical and numerical algorithm.

2. Mathematical Models

3D model for temperature distribution and electric field in metallic materials

The proposed 3D mathematical model consists of determining the temperature distribution \( T = T(x,t) \) and the electric field \( \bf{E} = \bf{E}(x,t) \) in the region
\[
D = \{ x = (x_1,x_2,x_3) \in \mathbb{R}^3 : x_i \in [0,L_i], i = 1,3 \} \quad \text{during any moment of time} \quad t \in [0,t_{\text{max}}] \quad \text{with the following equations as well as the initial and boundary conditions:}
\]

- dual-temperature heating dynamics for electron \((T_e(x,t))\) and lattice \((T_l(x,t))\) subsystem of metal materials (taking as an example copper) with femtosecond laser pulse:

\[
\begin{align*}
\rho_e \frac{\partial T_e}{\partial t} + \tau_e \left[ \frac{\partial^2 T_e}{\partial t^2} \right] &= \text{div}\left( \chi_e \left( x,t,T_e,T_l \right) \bf{E} \right) + F(x,t,T_l) = 0 \\
\end{align*}
\]

- electric field distribution on the periodically structured surface of the metallic material:

\[
\]
and the surface of the metal material at the initial and boundary calculated by the Beer-Lambert-

\[ \Delta_{\gamma_{\nu}} E_{\nu} - \sigma \left| \mathbf{E} \right|^2 - \]
\[ \left( \frac{\partial}{\partial t} E_{\nu} + \frac{\partial}{\partial \mathbf{x}} \frac{\partial}{\partial \mathbf{x}} E_{\nu} \right) \left( x, t \right) \in D \times [0, t_{\text{max}}], \]
\[ \forall (i, j, k) = 1, 3; j \neq k \neq i; \]

- the following initial conditions:

\[ T'(x, t)_{t=0} = T'_0 \left( x, x \in D, \right) \]
\[ \frac{\partial T'(x, t)}{\partial t} \bigg|_{t=0} = 0, x \in D, \]
\[ T'(x, t)_{t=0} = T'_0 \left( x, x \in D, \right) \]
\[ \frac{\partial T'(x, t)}{\partial t} \bigg|_{t=0} = 0, x \in D, \]
\[ \bar{E}(x, t)_{t=0} = \bar{E}_0 \left( x, x \in D, \right) \]
\[ \overline{\frac{\partial E(x, t)}{\partial t}} \bigg|_{t=0} = 0, x \in D; \]

- and the following boundary conditions:

\[ \left\{ T'(x, t) - k_1 \left( \frac{\partial T'(x, t)}{\partial x_i} \right) \right\}_{t=0} = T'_{i,0} \left( x/\{x_i\}, t, \right), \]
\[ x \in D, t \in [0, t_{\text{max}}], \forall i = 1, 3; \]
\[ \left\{ T'(x, t) - k_2 \left( \frac{\partial T'(x, t)}{\partial x_i} \right) \right\}_{t=0} = T'_{i,0} \left( x/\{x_i\}, t, \right), \]
\[ x \in D, t \in [0, t_{\text{max}}], \forall i = 1, 3; \]
\[ \left( \frac{\partial T'(x, t)}{\partial x_i} \right)_{x=L_i, t=0} = T'_{i,L_i} \left( x/\{x_i\}, t, \right), \]
\[ x \in D, t \in [0, t_{\text{max}}], \forall i = 1, 3; \]
\[ \left( \frac{\partial T'(x, t)}{\partial x_i} \right)_{x=L_i, t=0} = T'_{i,L_i} \left( x/\{x_i\}, t, \right), \]
\[ x \in D, t \in [0, t_{\text{max}}], \forall i = 1, 3; \]

- div\text{A} \equiv \sum_{i=1}^{3} \frac{\partial A_{i}}{\partial x_i};

- \nabla \text{A} \equiv \left( \frac{\partial A_{i}}{\partial x^1}, \frac{\partial A_{i}}{\partial x^2}, \frac{\partial A_{i}}{\partial x^3} \right)^{T};

- D denotes the interior field \( \overline{D} \), i.e.

\[ D \equiv \overline{D} \left( x = (x_1, x_2, x_3) \in \mathbb{R}^3, x_1 \in [0, L_1], i = 1, 3; \right); \]

- \( c^2_i \) and \( c_i \) are respectively the electron gas and heat capacities per unit volume \( V \);

- \( r \equiv 1 \) and \( k_i \) \( (i = 1, 2) \) are positive constants;

- the function \( \chi(x, t; T', \bar{T}) \) describes the electron thermal conductivity, which plays a pivotal role in the metals, semimetals, as well as certain semiconductors and superconductors. The electron thermal conductivity provides information on the carrier scattering mechanism, on the peculiarities of the band structure of the material, as well as information about the magnitude and temperature dependence of the bandgap (energy gap) of the material. It is important to note that the electron conductivity at low temperatures is determined by the scattering of electrons by impurities and defects and also increases with temperature. However, at high temperatures the process is determined by scattering on phonons and decreases with increasing temperature and therefore, at a certain temperature the electron thermal conductivity reaches a maximum, higher than that of a perfect crystal. The conductors in electron thermal conductivity are related to the conductivity of the Wiedemann-Franz Law; in superconductors electrons are combined into Cooper pairs, not involving in the transfer of heat. Furthermore, when the temperature falls below the transition temperature to the superconducting state, the electron thermal conductivity is determined by the normal (unpaired) electrons and decreases exponentially as it approaches zero. In bipolar semiconductors and semimetals, there is an additional mechanism called a bipolar component, namely, electron-hole pairs generated at the hot end of the target and diffusing towards the temperature gradient and recombinating at the cold end to produce heat;

- the function \( F(x, t; T') \), calculated by the Beer-Lambert-Bouguer Law:

\[ F(x, t; T') \equiv \left( 1 - \Phi \left( T' \right) \right) \cdot I \cdot e^{-\sum_{i=1}^{3} \beta_i \left( x, t, \right)} \times \right]_{t_{\text{laser pulse}}}^{t_{\text{laser pulse}}}[0, t_{\text{laser pulse}}]; \]

\[\text{describes the attenuation/absorption of laser radiation on the feeding surface of the target "flow" of free electrons due to the inverse scattering brake; in Equation (15) the function } \Phi, \text{ depending on the temperature } T', \text{ the surface of the metal material at the initial moment of time, is the reflectance of the target material; } I \text{ denotes the intensity of the incident laser beam on the target; } \beta_i \left( i = 1, 3 \right) \text{ is the absorption coefficient of the target material in the direction of the coordinate axis } \overline{OX}_i; \delta(\cdot) \text{ is the Dirac Delta Function; the function } \gamma(x, t) \text{ is the coefficient absorption in the resulting plasma laser flare; } t_{\text{laser pulse}} \quad \text{ denotes the end of the laser radiation; }\]
the scalar function $E_i(x,t)$ is the $i$-th ($i=1,3$) component (corresponding axis $Ox^i$) of the electric vector field $\vec{E}(x,t)$;

- $\varepsilon$ and $\mu$ mean respectively dielectric and magnetic permeability of the metal material (in our situation copper);
- $\sigma$ is a specific conductivity of the material;
- $c$ is the speed of light. Here it should be noted that for the simulation of the femtosecond laser radiation, the nested (embedded) laser energy distribution may exclude relief modulation over time $t_{\text{max}}$, however, it is crucial to consider that in the photoexcitation process, the copper surface for heating ultrafast electronic subsystem of the material to the temperature $T^r$ of the order of several electron volts, can significantly affect the optical properties of copper. It is easy to show at a specific wavelength (for example, $\lambda = 700$ nm) the femtosecond laser pulse by calculating the dependence $\Re(\varepsilon)$ and $\Im(\varepsilon)$ to $T^r$; consequently from the Drude Model considering the contribution of the lattice $\varepsilon'$ we can apply the following formula to determine the dielectric constant $\varepsilon'$:

$$\varepsilon = \varepsilon' - \frac{\omega_{\text{plasma}}^2}{\omega^2} e^{-\frac{1}{\omega_{\text{electron collision}}^2}} - e^{-\frac{1}{i \cdot \omega_{\text{electron collision}}^2}} + 1,$$

(16)

where $\omega$ is the frequency of the femtosecond laser pulses;

$$\omega_{\text{plasma}} = \frac{\rho_{\text{electron density}} e^2}{\sqrt{\varepsilon_0 m_{\text{effective mass}}}},$$

(17)

has a plasma frequency when the effective mass of the electron $m_{\text{effective mass}} \approx \frac{3}{2} m$ and the electron density $\rho_{\text{electron density}}$ for copper;

- $t_{\text{electron collision}}$ is a collision of electrons within the random phase approximation (see, for instance, [11]), and it is determined by the following formula:

$$t_{\text{electron collision}} = t_{\text{electron collision}} (T^r) \approx \frac{1}{K_{\text{electron-electron scattering}}} \times \left( 1 + \frac{h \omega}{T^r} \right)^2,$$

(18)

with $K_{\text{electron-electron scattering}}$, which for the investigation of the material (in our situation copper) can be determined by tabular dependence approximation $\varepsilon(\omega)$ at $T^r = 0$.

Thus, under certain assumptions the foregoing non-linear 3D model (1)-(18) describes the evolution of the temperature field and the electric field distribution on the surface of metals (taking as an example copper) under the action (irradiation) of femtosecond laser pulses. Since dielectrics with a forbidden gap in the beginning of laser radiation’s exposure there is no absorption of the radiation pulse due to reverse brake dissipation; hence, most dielectrics with relative moderate intensities of laser radiation can be considered as transparent. It is important to observe that the mechanism of interaction of femtosecond laser pulses with insulators is significantly different from the above-described mechanism for metals.

When the threshold of intensity, depending on the characteristics of the particular dielectric surface structure, begins to break down, the changing of the reflectance and absorbance of the surface is followed by the melting and ablation of the dielectrics (with a further increase in the intensity of radiation).

Following the outline of monograph [1] below, the main process leading to the synthesis of nano-sized particles is the influence of femtosecond laser pulses in the dielectric materials. At the initial stage of the action of a femtosecond laser pulse on a dielectric surface (we neglect the presence of impurities and defects in the original) the multiphoton absorption mechanism forms a free primeval / seeding electrons, and these electrons efficiently absorb laser energy by dissipating the reverse braking process, with energy $E^r$ beginning to increase rapidly and reaching the critical value, while developing process of ionization by electron impact (the so-called avalanche / cumulative ionization). For laser pulse durations of the order of hundreds of femtoseconds with a capacity in the range of $10^{10}-10^{14}$ W/cm$^2$, the tunnel ionization mechanism can be neglected. Then the dynamics of the electron density in the conduction band can be described by the following simplified equations:

$$\frac{\partial \rho^e}{\partial t} = \left( g^{ni} \right)^e \cdot I^n (t) + g^{ei} \cdot I^e (t) \cdot \rho^h;$$

(19)

$$\frac{\partial \rho^h}{\partial t} = \left( g^{ni} \right)^h \cdot I^n (t) + g^{eh} \cdot I^e (t) \cdot \rho^e,$$

where the superscripts $e$ and $h$ refer, respectively, to the electrons and the holes; $g^{ni}$ and $g^{ei}$ mean respectively multiphoton ionization rates (multiphoton ionization) and avalanche ionization (cumulative ionization); $I(t)$ indicates the intensity of the radiation incident on the target; and $n$ is the number of photons required for the excitation of electrons at the transition from the valence band to the conduction band.

The appearance of a large number of electrons in the conduction band of the dielectric leads to metallization of the dielectric surface layer. Therefore, it increases the absorption and reflection coefficients. However, unlike the metals, the density of electrons in the conduction band of dielectrics depends strongly on the time and distance target depth. Furthermore, it is possible to describe the electron- lattice relaxation dielectric process, by analogy with the description of metals. In addition, at the end of femtosecond laser pulses, with the passage of time, there are processes of recombination of electrons and three-body recombination (Auger recombination / three-body recombination), from which there is a gradual return of the target dielectric properties. Typically, after the irradiation in the process of recombination of electrons are formed various defective state whose lifetime depending on the type of the dielectric ranges from microseconds to days. Repeated action of femtosecond laser pulses lead to a more efficient heating by priority field defects with larger absorption cross section compared with the electrons of the valence band.

### 1D model for metallic materials

In conclusion, this paper presents the basic equations in 1D approximation describing the mechanism of action of femtosecond laser pulses in the dielectric materials based on photoemission and electric fields.

- The first equation is a modification (towards complexity – a detailed description) of the simple equations (19):

$$\frac{\partial \rho^e}{\partial t} = \theta \frac{\partial \rho^e}{\partial x} + \mu \cdot E \cdot \frac{\partial \rho^e}{\partial x} + \frac{\mu}{\varepsilon_0} \cdot E \cdot \rho^e - \rho^e \cdot \left[ \rho \right]^{\Delta} +$$

(20)

$$+ \left( g^{ni} + V^{ei} \right) \frac{\rho^{ni \Delta}}{\rho^{ni \Delta} + \rho^h} - F^{he} + F^{ph e};$$

- The Poisson’s equation:

$$\frac{\partial E}{\partial x} = -\frac{\left| \rho \right|}{\varepsilon_0 \cdot \varepsilon};$$

(21)
• The following is a formula for calculating the intensity of femtosecond laser radiation onto a target Gaussian time profile:

\[ I(t) = \frac{\ln 2}{\pi} \frac{2 \cdot f \cdot e^{-kx^2}}{t_f} \]  

(22)

The radiation intensity equation inside the dielectric depending on the time and distance of the target depth considering the multi-photon absorption and absorption by free electrons:

\[ \frac{\partial I(x,t)}{\partial x} + a(x,t) \cdot I(x,t) = -\frac{g_{ph}}{\rho^a + \rho^b} \]  

(23)

• The following formula calculates the absorption coefficient of a free electron:

\[ a(x,t) = 4 \cdot \pi \cdot k(x,t) \frac{\lambda}{\lambda} \]  

(24)

• The Fresnel’s formula for the estimation of the reflection coefficient:

\[ \Phi(t) = \left\{ \begin{array}{ll} \Re(e) + \sqrt{\Re(e)^2 + \Im(e)^2} & -1 + k^2 \\
\Re(e) + \sqrt{\Re(e)^2 + \Im(e)^2} & +1 + k^2 \end{array} \right. \]  

(25)

• The Einstein’s formula for the diffusion coefficient:

\[ \theta = \frac{k \cdot T^* \cdot \mu}{\varepsilon} \]  

(26)

• Formula estimation for the photoemission:

\[ \varepsilon_{ph} = \frac{(\rho_{ph}^a + \rho_{ph}^b) \cdot I(x,t)}{2} \]  

(27)

In the equations (20)-(27) there are the following designations (the designations that occurred earlier below are not listed):

• \( V^{\pm} = g_{ph} \cdot I \cdot \rho^+ \) the acceleration/development of cumulative ionization;

• \( g^+ = \frac{\rho^+}{\tau_i} \) the recombination velocity electron capture of the coupled state;

• \( \rho^+ \) the neutral atoms;

• \( \xi \) the characteristic depth of the target, which photoemission depth (maximum photoemission takes place on the surface of the dielectric material) decreases exponentially.

3. Conclusions

In the present paper, 3D model (1)-(18) as well as 1D model (only main equations without any initial or boundary conditions) (20)-(27) are proposed. The model (1)-(18) describes evolution of temperature field (scalar quantity) and electric field (vector quantity) in the case when femtosecond laser pulses act on the metallic materials; the model (20)-(27) contains the basic equations describing the simplified mechanism of action of femtosecond laser pulses on the dielectric materials.

In the future, the authors of this paper intend to: firstly, to develop an algorithm for finding a stable solution of the constructed 3D model (1)-(18); secondly, to compile the model (1)-(18) by adding other equations and conditions (which were not considered in this model) that can be occurred in the mechanism of femtosecond laser pulses action on the metallic materials; thirdly, to analytically solve the proposed 1D model and to carry out the deeply analyse of the obtained results.

4. Acknowledgements

For the first author the present article was executed within the framework of the State Research Programme "Next generation Information and Communication Technologies (NexIT)", Project No. 4.

References


AUTOWAVES OF LOCALIZED PLASTIC DEFORMATION ON THE YIELD PLATEAU AND ON THE WORK HARDENING STAGE

Prof. Dr. Phys.-Mat. Sci. Danilov V.1,2, Cand. Phys.-Mat. Sci. Gorbatenko V.1,2, Prof. Dr. Phys.-Mat. Sci., Zuev L.1,3
Institute of Strength Physics and Material Science Siberian Branch of Russian Academy of Sciences’,
Tomsk Polytechnic University1, Tomsk State University2 – Tomsk, Russia
dvi@ispms.tsc.ru, gvv@ispms.tsc.ru, lbz@ispms.tsc.ru

A study was made of the macro-scale plastic flow non-homogeneities, which occur in metals in the form of Luders bands or Portevin-Le Chatelier effect. The motion kinetics was investigated for the mobile fronts of Luders bands observed for the yield plateau as well as localized plasticity fronts traveling in the course of serrated plastic flow behavior (Portevin-Le Chatelier effect). It is shown that the propagation of the above two kinds of band fronts can be regarded as macro-scale auto-wave processes of switching and excitation, respectively, which frequently occur in active media of different kinds.

KEYWORDS: INSTABILITY OF PLASTIC FLOW, LUDERS BANDS, PORTEVIN-LE CHATELIER EFFECT, AUTOWAVES OF LOCALIZED PLASTIC DEFORMATION

1. Introduction

Investigations of the plastic deformation in solids have been carried on for 25 years. On the macro-, meso-and micro-scale levels the plastic flow is found to exhibit a non-uniform behavior from the yield limit to the fracture of material [1-4]. In the course of plastic flow development a changeover in the modes of localized plasticity auto-waves would occur, with each mode corresponding to the respective flow stage on the loading curve plotted for deforming material. Thus the phase auto-waves would form at the stage of linear hardening; a stationary dissipative structure would emerge at the stage of parabolic hardening (Tailor stage); collapse of the auto-wave would occur at the pre-fracture stage where the onset of necking takes place [4]. The phase auto-wave is represented by a set of equidistant localized deformation nuclei moving in a concerted manner in one and the same direction. The phase auto-waves have been considered in greater detail; the spatial period (wave length) and propagation rate can be determined for this kind of auto-waves [4-6]. In a particular case, the number of moving nuclei might be no more than four; hence, the motion of these nuclei would exhibit no spatial periodicity. However, a correlation has been established for the motion rates of nuclei propagating at the yield plateau. The latter stage is characterized by plastic flow instability, which is generally referred to as Luders band (LB) propagation [7]. The plastic deformation is also known to exhibit another type of space-time non-homogeneity, the so-called effect of Portevin-Le Chatelier (PLC) or serrated flow, which would change the traditional shape of the deformation curve due to the appearance of peaks [8]. The both types of plastic flow instabilities and their formation and development in the deforming medium are addressed herein in the types of plastic flow instabilities and their formation and development in the deforming medium are addressed herein in the

2. Materials and experimental procedures

The LB propagation was studied for the test samples of low carbon hot rolled steel, which had near ferrite structure and average grain size ~20 μm (steel composition 0.05…0.11 % C; 0.35…0.65 % Mn; 0.05…0.17 % Si). The PLC effect was studied for naturally aged duralumin containing Al-based solid solution having grain size ~30 μm and hardening particles having submicron size (duralumin composition 3.8…4.8 % Cu; 0.4…0.8 % Mn; 0.4…0.8 % Mg). Note: alloy composition is in mass percent. The test samples having dog-bone shape were cut out of steel plates; the sample gauge had dimensions 50×10×3 mm. The samples were tested in tensile loading in the testing machine Walter+Bai AG LFM-125 at the rate \( v_{\text{max}} = 3.3 \times 10^{-4} \text{ m/s} \) (deformation rate \( \dot{\varepsilon} = 6.6 \times 10^{-5} \text{ s}^{-1} \)) and at the temperature 300 K.

In conformity with the requirements of the experimental technique, the test samples had flat diffuse reflecting gauge. Using the method of digital statistical speckle photography, the visualization and nuclei kinetics registration was performed for the localized plasticity zones occurring on the sample surface [10]. The sample was illuminated by the coherent light of a semi-conductor laser having wavelength 635 nm and power 15 mW. The speckle pattern superimposed on the respective image was registered, digitized with the aid of video camera PixeLink PL-B781 having frequency 10 Hz and then stored. A sequence of counts was formed for each image point, which characterized variation of image brightness with time. Then the dispersion and mathematical expectancy were calculated; the ratio of the values obtained was used for mapping deformation localization zones. The above technique makes feasible in situ registration of material zones in which deformation localization occurs for a given increment in tensile sample length. It can be seen that the localized deformation zones show up in the video images as bright bands.

2.1. Development of Luders bands

It was found that the micro-plasticity stage will be invariably realized in the deforming samples of low carbon steels [11]. Therefore, the registration of localized plasticity zones was started before the upper yield limit was attained (Fig. 1).

![Fig. 1. The yield plateau on the deformation curve plotted for the test sample of low carbon steel](image)

Speckle image recording was terminated as soon as the yield plateau was over and the deformation hardening stage began. It was established experimentally that the plastic deformation would first localize in the form of LB nucleus emerging at the micro-plasticity stage. The appearance of LB nucleus might be traced from acoustic emission signals [12]. As is seen from Fig. 2, a narrow wedge of deformed material would advance crosswise through the test sample at the rate \( \dot{V}_{\text{max}} \approx (0.4\ldots1.2) \times 10^{-2} \text{ m/s} \). The diagram \( \sigma(\varepsilon) \) represents the nucleus growth as a fairly sharp yield point having an ascending and a descending branch (Fig. 1). As soon as the LB nucleus traverses the sample cross-section, the formation of LB is completed and its widening would begin (band widening observed in the diagram \( \sigma(\varepsilon) \) corresponds to the yield plateau). Now that the
LB formation is over, one front would emerge at the LB foreground and the other at the rear. The both fronts start traveling in the opposite directions along the sample axis at rate ±V_f (Fig. 3). Sometimes a LB band and its two fronts would form at the clamp of the testing machine; then one of the fronts would almost immediately leave the clamp and become immobilized.

![Image 2](image2.png)

**Fig. 2.** The nucleation of LB; the images were obtained at time intervals of 7 s

The motion rate of LB fronts is an important characteristic of the process. As soon as a single LB appears (see Fig. 3), two of its fronts start moving in opposite directions at practically the same rates, i.e. \( V_f \approx \pm 8 \times 10^{-5} \) m/s; hence, \( V_f \approx 0.1 \ V_{nucl} \). As one of the fronts approaches the clamp of the testing machine, its rate drops down to zero, while the rate of the other front increases up to \((1.3…1.5) \times 10^{-4} \) m/s.  

![Image 3](image3.png)

**Fig. 3.** The ingrowth and propagation of LB; the images were obtained at time intervals of 60 s

When a couple of LBs forms simultaneously, four fronts start traveling at practically the same rate \((3…5) \times 10^{-5} \) m/s, which is somewhat lower than in the former case (see Fig. 4). As soon as fronts 1 and 4 approach the testing machine clamp, their rate drops off and, both fronts would finally become arrested. Fronds 2 and 3 continue moving in the opposite directions at the rate which is almost twice that of the original rate. The fronts of two neighboring LBs would become annihilated on meeting. The original fronts are inclined to the extension axis at an angle \( \approx \pi/3 \); however, in the course of front motion towards the clamp the inclination would drop off to \( \approx \pi/2 \).

![Image 4](image4.png)

**Fig. 4.** The propagation of two LBs; the images were obtained at time intervals of 60 s

Thus in the case of mobile LB front, which emerges in the deforming sample at the yield plateau at the pre-assigned value \( M_{def} \), the following rule holds true:

\[
\sum_{i=1}^{N} |V_f^i| = \langle V_f \rangle = \text{const} ,
\]

(1)

where \( |V_f^i| \) is the modulus of the motion rate of \( i \)-th LB front; \( N \) is the number of fronts traveling in a concerted way and \( \langle V_f \rangle = (1.6 \times 10^{-4} \) m/s is the generalized rate of growth of the plastically deforming zone, which emerges at the yield plateau in the test sample. The relationship between the velocities \( V_{mach} \) and \( \langle V_f \rangle \) is found from the equality of the time it takes for the given plasticity zone to cover sample gauge length, \( L \), and of the time it takes for a given sample to acquire ultimate elongation at the yield plateau, \( \delta L \).

\[
\frac{L}{\langle V_f \rangle} = \frac{\delta L}{V_{mach}} .
\]

(2)

The above suggests that such is indeed the case: the generalized growth rate of deformed zone area is by about one order greater than the rate of the movable clamp of the testing machine, i.e.

\[
\langle V_f \rangle = \frac{L}{\delta L/v_{mach}} > V_{mach}.
\]

(3)

2.2. Development of serrated flow

The PLC effect is manifested in duralumin at room temperature in the course of deformation up to necking (see Fig. 5).

![Image 5](image5.png)

**Fig. 5.** The serrated flow observed for the test sample of duralumin; the insert shows toothing observed for the studied range of stresses

It can be seen from the diagram \( \sigma(\varepsilon) \) that the intermittent flow is characterized by the appearance of type A, B and C toothing, with the type of toothing varying with sample elongation [14]. Thus the occurrence of a C type tooth corresponds to a single sharp decrease in the deformation, while the appearance of A type tooth is suggestive of working stress drops alternating with constant or weak increase in the stress. The deformation diagram was obtained for the linear work hardening stage in the test sample of duralumin for the strain range \( 0.01 \leq \varepsilon \leq 0.03 \). It shows mostly C type toothing, which can be attributed to separate localized deformation bands corresponding, in terms of morphology, to non-complex material areas. The PLC bands will originate in the middle of the sample gauge or on the lateral side of the sample in the vicinity of the clamp of the testing machine. The bands occurring in the tensile sample would expand throughout the sample cross-section. The rate of growth of PLC nuclei is \( \approx 1 \) m/s, which is three times that of the LB nuclei. The band having an ingrown nucleus has a forefront and a rear-front with \( \approx 2…3 \) mm separation in between (Fig. 6); it is inclined to the extension axis at an angle \( \approx \pi/4 \). The band would travel along the sample to the testing machine clamp at a rate \( \pm V_{ab} \).
The modulus of band propagation rate is $1.3 \times 10^{-3} \leq V_\lambda \leq 1.8 \times 10^{-3}$ m/s, which is ten times that of the PLC fronts' propagation rate or even higher. A sharp drop of the working stress is observed in the diagram $\sigma(\varepsilon)$; it corresponds to the process of nucleus ingrowth, while the motion of the PLC band (single C-type tooth) corresponds to an increase in the working stress up to the nominal value. As soon as the PLC band reaches the clamp of the testing machine, a drop in the working stress would occur, with resultant emergence of a new band in the middle of the sample. The new band would start traveling along the extension axis in the opposite direction relative the former band (see Fig. 6). As soon as the new band reaches the clamp of the testing machine, the same process is repeated once again. Thus, the serrated flow behavior has a characteristic feature: the PLC bands of deformation localization would travel repeatedly along the sample gauge, with each run of a PLC band over half the gauge length corresponding to a single C-type tooth in the diagram $\sigma(\varepsilon)$. The same periodic motion of PLC band fronts would occur at the stage of parabolic deformation hardening; however, the motion rate of fronts would drop off gradually and at the onset of necking the fronts would become immobilized.

### 3. Discussion of results. Matching of the localized deformation behavior at the LB fronts and the PLC effect

The studies were performed on the base of experimental evidence. The kinetics of mobile LB and PLC fronts suggests that they have both similar and dissimilar features. Thus the plastic deformation tends to localize within narrow zones of material; hence, the both types of band fronts are macro-scale manifestations of this tendency. In both cases the kinetics of band emergence is similar, since localized plasticity nuclei would form on the lateral surface of the sample gauge to grow finally through the sample cross-section. As soon as the ingrowth of nuclei is completed, a sharp distinction between the two kinds of processes becomes manifest. Thus the evolution of LB is in point of fact its widening within the sample gauge, with the entire plastic deformation becoming localized within the mobile fronts traveling in opposite directions, with their intricate shape varying in the process. The LB fronts would move independently of one another; however, their motion rates would vary in such a way that the sum of moduli of the fronts’ rates remains constant (Eqn. 1). In case two or more LBs start broadening, their fronts are liable to annihilate on meeting.

Due to the PLC effect, the plastic flow would become localized within a single band, which excludes the possibility that bands would annihilate on meeting. We are reminded that the LB fronts would propagate over the sample gauge only once, while the PLC front would travel repeatedly along the sample gauge. This is an added reason to state that the effects in question differ significantly.

The above distinctions are considered in the frame of auto-wave model of plastically deforming medium by the workers in [4]. One of the main assumptions of the model is that the deforming medium is an active one; hence, it must contain energy sources distributed in the material volume [13]. By addressing the problem of plasticity, it might be reasonable to consider elastic stress carriers as energy sources, which are capable of relaxation when crystal lattice defects are formed in the deforming solid.

In the frame of well-known microscopic models of LBs [6, 10], the medium within the mobile front would pass from an elastically stressed state to a plastically deformed state due to the relaxation of stress carriers. The both states are stable ones; however, the elastically stressed state is a metastable one and the plastically deformed state, an absolutely stable one; hence, no transition is feasible from the latter state back to the former. In point of fact, the microscopic models of LB have nothing to do with the auto-wave theory [13]. However, the relaxation of stress concentrators occurring in the medium enables one to place the given medium into the category of trigger type (bistable) systems. Hence, the perturbation front might be regarded as a switching auto-wave. The acts involving a changeover in the medium’s states are irreversible ones, which precludes repeated passing of perturbation fronts over one and the same area. For this reason, two switching auto-waves traveling in opposite directions would generally annihilate on meeting, which is the case with the LB fronts.

The PLC effect and the micro-mechanisms involved have been studied in detail [7, 12]. The results suggest that the components of the active medium might be in a different sequence of states. At the PLC front the components may also pass from the metastable elastically stressed state to the plastically deformed state. However, the latter state is never absolutely stable, since the PLC effect development occurs at the hardening stage. As the stress level increases up to the nominal value, which is manifested by the formation of C type tooth, the active elements of the deforming medium will return to the original metastable elastically stressed state. In conformity with the theory of auto-wave processes, such elements and such active media is excitable to occur.

The excitable element is modeled by a three-state cycle. Thus an excitable element, which is fully recovered from any previous excitation, is in state A (at rest); state B is the excited state of the excitable element and state C is the refractory state of the excitable element. By way of illustration, let us consider the stress carrier as an excitable element. State A is a meta-stable elastically stressed state of the stress carrier; state B is the gradual disappearance of stresses from the stress carrier and state C is the refractory state of the stress carrier with the stresses increasing. A perturbation will excite the quiescent stress carrier, which would become excited at random times following an external stimulus, e.g. when subjected to external periodic forcing by another excitable element or due to thermal fluctuations. The transition from a refractory state to rest takes place in the excitable element at the hardening stage due to the steadily increasing level of external stresses. So long as the stress carrier is in a refractory state, it would remain inactive. As the localized deformation front travels along the sample, all the stress carriers would become excited following an external stimulus and the medium would acquire a refractory state. The stress concentrators would pass into the state of rest after the lapse of refractory time; then another deformation front would propagate in the sample. The above events summate to produce the macroscopic dynamics of the whole system [13]. In view of the above, the PLC effect is attributed to the propagation of excitation auto-waves in the deforming medium.

### 4. Conclusions

1. It is shown that the LBs and the deformation bands related to the PLC effect are initiated in the deforming medium via one and the same mechanism, i.e. nucleus ingrowth throughout the sample section on the lateral sample side.
2. It is found that the rate of deformation bands which form due to the PLC effect is three times that of LB.
3. LB fronts and deformation bands related to the PLC effect are switching and excitation auto-waves, respectively, which form in deforming active media.

The study was performed in the frame of fundamental scientific research program of Russian Federation state academies of science for the period 2013–2020 with partial support of the RFBR Grant.
The experimental studies were conducted on the equipment made available by the courtesy of the Nanotech Center at the ISPMS SB RAS.

5. References


MATERIALS FOR ELECTROMAGNETIC INTERFERENCE SHIELDING

Ass.prof. PhD Kamelia Ruskova¹, Assos. Prof. PhD Ljudmila Taneva², Assos. Prof. PhD Alexandar Lirkov¹
Technical University, Sofia, Bulgaria ¹, South-West University “Neofit Rilski”, Blagoevgrad, Bulgaria ²

e-mail: kruskova@tu-sofia.bg

Abstract: The electronic devices are everywhere in our life and for their proper performance it is very important to have them protected from Electromagnetic Interference (EMI). Sources of EMI could be some components of the device, nearby conductors, other devices, high voltage lines, atmosphere static discharges, etc. One of the main methods for protection from external EMI is by shielding the whole device and for this purpose are used various materials. In this paper are presented two types of materials used for EMI shielding - polyester silk with electrolysis metal coating and nitrile butadiene based compounds (NBR) containing as a filler natural magnetite. For both materials is investigated the EMI absorption for electromagnetic waves in the band (8-12 GHz) and at 2380 MHz.

Keywords: ELECTROMAGNETIC INTERFERENCE, SHIELDING, FILLED ELASTOMERS, PLASTIC ELECTROMAGNETIC INTERFERENCE SHIELDING

1. Introduction.

Shielding from EMI is a complex process associated with propagation of electromagnetic waves in environments with various electromagnetic properties and their interaction with the object, characterized with reflection, refraction, scattering and absorption. For quantitative evaluation of the effectiveness of the protection is used shielding coefficient (S) equal to the ratio of the amplitude of the intensity of the electric or magnetic field at any point in the shielded area compared to the amplitude of the intensity of the field at the same point in the absence of the shield:

\[
S = \frac{E_o}{E_o} = \frac{H_o}{H_o}
\]

When an electromagnetic wave reaches the shield, one part of it is reflected and another penetrates it undergoing absorption and multiple internal reflections (MIR) as legitimately some portion of it passes through the shield. All processes of reflection and absorption lead to loss of electromagnetic energy:

\[
A(\text{dB}) = A(\text{refl}) + A(\text{abs}) + A(\text{MIR})
\]

The energy loss from absorption (absorption loss) does not depend on the type of the electromagnetic waves (electric or magnetic field) and goes up with increasing the waves frequency and the thickness of the shield, as well as with increasing the magnetic permeability of the material from which it is made. [1] The energy loss from reflection (reflection loss) depends on the ratio of the impedance of the waves and the impedance of the shield material. It is larger for electric fields, decreases with increasing waves frequency, and does not depend on the thickness of the shielding material. Multiple internal reflections always reduce the shielding effect because each reflection inside the screen allows a small amount of energy to pass through it.

2. Shielding methods.

Electromagnetic interference (EMI) is emitted by every wire or electronic device associated with electrical energy transferring, sun radiation, high voltage atmosphere discharge and other sources. The rapid development of the electronic industry during the second half of XX century resulted in implementation of electronic devices everywhere. Each of them is a source of EMI, jeopardizing the normal functionality of the electronic devices, thus bringing out in the foreground this problem.

Basically the protection from undesired signals emitted from electronic devices could be achieved by three methods: proper design; electromagnetic shielding of separate device modules; shielding of the whole device. The last method is the easiest for realization and economically profitable especially because of the possibility for plastic shielding realization. The frequency interval above 10 GHz is radio frequency and is the most common in the everyday life.

For electromagnetic shielding are used various materials - metal foils, metal grid, cloth with metal knitted fibers, materials with electro conductivity fibers and absorbents - wood pulp, micro porous rubber, polyvinylchloride resins with fillers, silicon rubber, graphite filled textolite, etc. (tab.1)

Table 1: Characteristics of some shielding methods

<table>
<thead>
<tr>
<th>Method</th>
<th>Width, µm</th>
<th>Resistance, Ω/ft</th>
<th>Shielding, dB</th>
</tr>
</thead>
<tbody>
<tr>
<td>Arc sputtering of zinc</td>
<td>15-25</td>
<td>0,03</td>
<td>50-60</td>
</tr>
<tr>
<td>Spraying of molten zinc</td>
<td>25</td>
<td>4,0</td>
<td>50-60</td>
</tr>
<tr>
<td>Acrylic paint with nickel powder</td>
<td>50</td>
<td>0,5-2,0</td>
<td>30-75</td>
</tr>
<tr>
<td>Acrylic paint with silver powder</td>
<td>25</td>
<td>0,04-0,1</td>
<td>60-70</td>
</tr>
<tr>
<td>Acrylic paint with copper powder</td>
<td>25</td>
<td>0,5</td>
<td>60-70</td>
</tr>
<tr>
<td>Paint with graphite filler</td>
<td>25</td>
<td>7,5-20</td>
<td>20-40</td>
</tr>
<tr>
<td>Cathode sputtering</td>
<td>0,75</td>
<td>1,5</td>
<td>70-90</td>
</tr>
<tr>
<td>Galvanic metalizing</td>
<td>0,75</td>
<td>0,1</td>
<td>85</td>
</tr>
<tr>
<td>Chemical metalization /copper-nickel/</td>
<td>1,25</td>
<td>0,03</td>
<td>60-70</td>
</tr>
<tr>
<td>Chemical metalization /silver/</td>
<td>1,25</td>
<td>0,5</td>
<td>70-90</td>
</tr>
<tr>
<td>Vacuum metalizing</td>
<td>1,25</td>
<td>5-10</td>
<td>50-70</td>
</tr>
<tr>
<td>Ion deposition of copper</td>
<td>1,0</td>
<td>0,01</td>
<td>50</td>
</tr>
<tr>
<td>Conductive plastic /nylon containing 40% carbon/</td>
<td>-</td>
<td>75-100</td>
<td>40-60</td>
</tr>
</tbody>
</table>
Nowadays microwave absorbers are mainly used for:
- protection of the environment and people from the adverse influence of high frequency electromagnetic radiation due to industrial and other sources;
- technical purposes like eliminating undesired signals and noise in radio and television technique; in anechoic chambers; antiradar camouflage of mobile and fixed military structure and objects for reducing or altering the radar signature.

The main microwave absorbers’ requirements are minimum reflection and maximum absorption of the electromagnetic energy. The ideal absorber has broad frequency range of the absorbing energy, excellent weather stability, light weight, characteristic reliability and capability to work in a broad temperature range.

3. Measurement of the attenuation coefficient of metalized fabrics

The measurement of the attenuation or shielding effect of the metalized fabric was conducted within 8-12 GHz band. The samples were installed between two coaxial waveguides for this frequency range - R100 (3x10mm) and pressed with equal force. The signal was transmitted from sweep generator HP 8620C and its changes were monitored on spectrum analyzer HP 8559A (both sets - Hewlett Packard make). The fabric samples size is 50mmX50mm. They are pressed with equal force between the two waveguides fully covering their openings.

The attenuation of the tested sample is calculated as follow:

\[ A = 10 \log P_{\text{pass}} / P_{\text{out}} \]

where

\[ P_{\text{pass}} - \text{power of the electromagnetic waves that passed through the shield; } P_{\text{out}} - \text{power of the electromagnetic wave outside the shield; } P_{\text{out}} = 4 - 5 \text{mW}. \]

4. Experimental results.

The paper presents two types of shielding materials. The first one is made of 100% polyester silk (K167 dtex.) with electrolysis metal coating - “sandwich” type construction - Cu/Ni-P [1].

Data was collected according BDS – specific area mass 84 g/m², breaking strength on the basis 74,7 daN, weft-72,2 daN, extensibility of the base and by weft - 400 n/10 cm, filling 77%, coefficient of drape 23%. With appropriate combination of thickness EMI attenuation above 50 dB is achieved in the band of the civil radiolocation (8-12 GHz) and at 2380 MHz. The received data of the experiments is given on tab.2, tab.3, tab.4 and tab.5.

Table 2: Electrical resistance of a square (Ω/10cm) depending on the conditions of the chemical metallization

<table>
<thead>
<tr>
<th>Chemical metallization</th>
<th>Direction of the measurement</th>
<th>10</th>
<th>20</th>
<th>30</th>
<th>45</th>
<th>60</th>
</tr>
</thead>
<tbody>
<tr>
<td>Coppering</td>
<td>by base</td>
<td>0,26</td>
<td>0,09</td>
<td>0,075</td>
<td>0,05</td>
<td>0,03</td>
</tr>
<tr>
<td></td>
<td>by weft</td>
<td>0,24</td>
<td>0,12</td>
<td>0,08</td>
<td>0,055</td>
<td>0,045</td>
</tr>
<tr>
<td>Coppering</td>
<td>by base</td>
<td>0,36</td>
<td>0,06</td>
<td>0,045</td>
<td>0,03</td>
<td>0,03</td>
</tr>
<tr>
<td></td>
<td>by weft</td>
<td>0,56</td>
<td>0,07</td>
<td>0,06</td>
<td>0,04</td>
<td>0,04</td>
</tr>
</tbody>
</table>

Table 3: Shielding effectiveness (attenuation, dB) depending on the chemical metallization type. Signal level 10mW, 10GHz

<table>
<thead>
<tr>
<th>Chemical metallization</th>
<th>Duration (min)</th>
<th>10</th>
<th>20</th>
<th>30</th>
<th>45</th>
<th>60</th>
</tr>
</thead>
<tbody>
<tr>
<td>Coppering</td>
<td>42</td>
<td>62</td>
<td>65</td>
<td>67</td>
<td>67.5</td>
<td></td>
</tr>
<tr>
<td>Coppering +10min nickel-plating /Ni-P/</td>
<td>57</td>
<td>65</td>
<td>66</td>
<td>71</td>
<td>72</td>
<td></td>
</tr>
</tbody>
</table>

Table 4: Shielding effectiveness (dB) depending on the duration of the nickel-plating of fabric A

<table>
<thead>
<tr>
<th>Electrolyte for nickel-plating</th>
<th>Frequency, GHz</th>
<th>5min</th>
<th>10min</th>
<th>20min</th>
<th>30min</th>
</tr>
</thead>
<tbody>
<tr>
<td>alkaline</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>pH=9,0</td>
<td>8,06</td>
<td>29</td>
<td>39,5</td>
<td>51,4</td>
<td>55,5</td>
</tr>
<tr>
<td></td>
<td>9,46</td>
<td>26,1</td>
<td>38,1</td>
<td>47,8</td>
<td>52,3</td>
</tr>
<tr>
<td></td>
<td>12,06</td>
<td>23,0</td>
<td>36,0</td>
<td>44,5</td>
<td>49,3</td>
</tr>
<tr>
<td>acid</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>pH=5,0</td>
<td>8,06</td>
<td>38,5</td>
<td>47,6</td>
<td>58,6</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>9,46</td>
<td>36,8</td>
<td>43,8</td>
<td>49,6</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>12,06</td>
<td>36,1</td>
<td>40,2</td>
<td>46,8</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 5: Shielding effectiveness (dB) depending on the duration of nickel-plating on fabric B (calendered)

<table>
<thead>
<tr>
<th>Electrolyte for nickel-plating</th>
<th>Frequency, GHz</th>
<th>5min</th>
<th>10min</th>
<th>20min</th>
<th>30min</th>
</tr>
</thead>
<tbody>
<tr>
<td>alkaline</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>pH=9,0</td>
<td>8,06</td>
<td>25,3</td>
<td>35</td>
<td>49,3</td>
<td>51,2</td>
</tr>
<tr>
<td></td>
<td>9,46</td>
<td>24,3</td>
<td>33,8</td>
<td>46,7</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>12,06</td>
<td>24</td>
<td>33,6</td>
<td>46,8</td>
<td>-</td>
</tr>
</tbody>
</table>

There were also investigated the properties of thin elastomeric microwave absorbers. The samples were built on the basis of nitrile rubber, containing natural magnetite (particle size below 10µ) as absorption active filler.

The mixing process of the rubber with the filler was carried out by Brabender Plasticorder at 50 °C. The rubber-filler ratio was 10-60 phr. The films (average thickness of 100 µ) were prepared from 2mass% 1,2-dichloroethane solution of the rubber-filler composition. Magnetic field with induction of 0,6T was applied for the period of 20 minutes during the film formation process. The magnetic field induction was controlled by teslameter. The samples were placed parallel and perpendicularly to the magnetic field lines of force.

The optical microscopy pictures give us information about the structure of non modified magnetite filled samples and another one modified with EMF. (Fig.1.) In presence of magnetic field we can
see an organization in filler particles ordering, which increases the probability the propagating wave to meet on its way a filler particle instead of polymer matrix material. This probability in the absence of EMF, when the structure has the usual disordered nature and the filler particles are situated chaotic is considerably less.

Fig.1: Optical microscope pictures (x 70) of nitrile rubber films, containing different concentrations of magnetite (H=0, without magnetic field application; H>0- with magnetic field application)

a) 10 phr Fe₃O₄, H=0
b) 10 phr Fe₃O₄, H>0
c) 50 phr Fe₃O₄, H=0
d) 50 phr Fe₃O₄, H>0

Fig. 2 shows the absorption of electromagnetic waves in the range 8-12 GHz for our samples. The effects concern the influence of the magnetic field on the magnetite distribution in the rubber matrix. The material is magnetically soft and its particles orientation is highly dependent on external magnetic fields.

The measurements were performed in a waveguide line. There were three measured absorbing samples - one control and two magnetically modified. For the first modified sample the direction of the external magnetic field is parallel with the wave in the waveguide. For the second sample it’s perpendicular. The absorption of the no modified sample is the lowest. The sample modified with a perpendicular magnetic field has the highest absorption.

Fig. 2. Absorption of electromagnetic waves in the 8-12 GHz range by films based on nitrile rubber and natural magnetite filler.

The change in absorption of the electromagnetic waves of the elastomeric microwave absorbers under the influence of an external magnetic field is due to the orientation phenomena in the elastomeric matrix, as well as the arrangement of the particles of the absorption-active filler in a favorable manner for better interaction between it and the electromagnetic waves falling on the shield [2].

5. Conclusions.

The researched materials were tested for effectiveness of the shielding effect. The attenuation was measured after application of magnetic field to the elastomeric microwave absorbents in both ways of operation of the electromagnetic field - perpendicularly and parallel to the sample in the range 8-12 GHz. The experiments proved that the tested materials have better shielding effect, compared to the control sample. For both materials the absorption in both directions was higher for the whole range 8-12GHz than the one of the sample. The observed improvement of the shielding effect is due to the interaction of the electromagnetic waves and the filler-magnetite. The oriented arrangement and resulting structure of the particles of the active compound under EMF influence leads to increase of the effect of shielding [3,4].


The application of magnetic field during sample preparation affects the filler particles ordering causing secondary structures formation guided by the direction of the magnetic field lines of force.
NEW CLASS OF COMPOSITE BRONZE, ARMED WITH STEEL DENDRITES FOR ANTIFRICTION TECHNIQUE

Prof. Dr. Potekhin B. A.1, Dr. Khristolyubov, A. S1, Prof. Dr. Hernández Fereira A.A.2
Ural State Polytechnic University, Ekaterinburg, Russia 1,
Universidad de Cienfuegos “Carlos Rafael Rodríguez”, Cienfuegos, Cuba 2
pba-nn@yandex.ru, archdez@ucf.edu.cu

Abstract: Bronzes type BrZhNA 12-7-1 (12% Fe, 7% Ni, 1% Al) form dendritic structure that consists of a maraging martensitic steel 03N15Yu1 mesh, which arms matrix given bronze with content Cu – base, 5-6% Ni and 1% Al. Dendrites can be strengthened by dispersion hardening. Mass transfer phenomena during crystallization and heat treatment in these bronzes were investigated. This bronze has elevated mechanical and technological properties, low friction coefficient and high wear resistance level during sliding friction.

Keywords: BRONZE, CRYSTALLIZATION, DENDRITE, COMPOSITE, MASS TRANSFER, MECHANIC PROPERTIES, FRICTION COEFFICIENT, WEAR RESISTANCE

1. Introduction

The widely known bronze BrO10 has high tribological, good casting and mechanical properties, but it is not deformed in the hot state due to the presence in it of brittle intermetallic Cu3Sn unfavorable form, that do not make possible the control of their properties. Other bronze such as BrA1, BrB2 which have a complex of high mechanical and technological properties, are not suitable for using in the sliding friction assemblies, since they do not meet the requirements of Charpy’s rule as they lack of a solid “supporting surface". Regarding this situation, the possibility of creation of bronze, reinforced with maraging steel dendrites of composition 03N15Yu1, whose properties are well known, was considered.

The present work is a continuation and development of researches [1-4] conducted on a group of composite bronzes, differing mainly in the methods of their manufacture, which showed that the production of castings rod (Ø=7 mm, l= 800 mm) by vacuum suction guarantee a speed of crystallization (> 700°C/s) that it is not lower than the crystallization rate from the melted metal. Essentially, it was carried out an experimental modelling of melted metals BrZhNA 12-7-1, BrZhNOA 12-7-2-1 in its pure (ideal) form, in the properties of which, do not affect the technology of welding or casting, and it was examined the structure of the studied alloys and welded joints bronze-bronze, bronze-cast iron and the mechanical and tribological properties of these bronzes.

The purpose of this work - the creation of bronzes, designed for sliding friction assemblies, with an optimal set of mechanical, technological and service properties.

2. Materials and Methods

Experimental bronzes ingots, weighing 3 kg were prepared by melting the pure raw materials in a Tamman furnace in a reducing atmosphere of carbon monoxide in alundum crucible. Ingots, 3 and 7 mm diameter rods, were made of bronze obtained by vacuum suction [5-8] from the melted metal, prepared in the induction furnace (Model API-25), in 3 and 7 mm diameter and 800 mm long quartz tubes. The particular feature of obtaining rods was that the suctioned melted bronze cooled in less than 1s from 1300 to ≥ 600°C (to dark cherry color) in a quartz tube. Crystallization and cooling rate in this temperature range was not less than 700°C/s. In fact, it can be called quenching from the liquid state similar to single-layer casting. Also, the ingots were melted again in argon using the procedure described in detail in [4]. The chemical composition of bronze studied is presented in Table 1.

<table>
<thead>
<tr>
<th>№ Composition</th>
<th>Bronze brands</th>
<th>Fe</th>
<th>Ni</th>
<th>Sn</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>1 BrZhNA 13-6-1</td>
<td>12,88</td>
<td>5,98</td>
<td>-</td>
<td>1,04</td>
<td></td>
</tr>
<tr>
<td>2 BrZhNOA 13-6-2-1</td>
<td>13,5</td>
<td>7,00</td>
<td>1,6</td>
<td>0,98</td>
<td></td>
</tr>
<tr>
<td>3 BrZhNA 12-7-1</td>
<td>11,85</td>
<td>7,43</td>
<td>-</td>
<td>0,48</td>
<td></td>
</tr>
<tr>
<td>4 BrZhNOA 12-7-2-1</td>
<td>11,30</td>
<td>6,96</td>
<td>2,13</td>
<td>0,79</td>
<td></td>
</tr>
</tbody>
</table>

Notes. 1) Total amount of impurities (Mn, Cr, Si and others) did not exceed 0.5 wt.%. 2) bronzes 1, 2 obtained in ingots, bronzes 3, 4 vacuum suction (vacuum casting).

Local chemical analysis of the structural components of the alloys was performed on a scanning electron microscope Jeol JSM 6490-LV with an attachment for microanalysis Oxford Inca Dry Cool (resolution of 133 eV) with a 3 µm diameter area. Total chemical analysis of alloys was determined with an area of 2 mm² by averaging the results of the three measurements from different sites.

Metallographic studies were carried out on the optical (Carl Zeiss Axio Observer) and scanning (Carl Zeiss EVO50) microscopes.

Alloy mechanical characteristics were determined using standard tensile test of tenfold specimens with a gage length diameter 5 ± 0,02 mm at room temperature in an Instron 3382 machine.

Friction coefficient and wear intensity were determined in a special device, mounted on the base of a CNC machine. Testing was carried out by a disk-finger specimen scheme with continuous computer recording of all test parameters (pressure, sliding speed, temperature, friction force). For each alloy were tested three specimens (finger samples of bronze different composition) of size 6 x 6 x12 mm; the counter body was a disk of steel ShKh15 (45 HRC). Tests have a comparative character, that is, the friction coefficient of the alloys was evaluated by comparing with the friction coefficient of the base material, which was a bronze BrO10. Over technique tribological test is described in detail in [3, 9].

Wear was determined by fingerprint method at a pressure of p=1 MPa and sliding speed v = 3,3 m/s in a friction path of 105 m. The wear rate was evaluated as the ratio of wear to the friction path.

Also were investigated two types of melting manufacture: bronze BrZhNA 12.7.1 melted on a BrZhNOA 12-7-2-1 ingot weighing 2 kg (cladding thickness up to 5 mm) and the same bronze melted on a Ø=30 mm rod made of cast iron SCh28, cladding thickness of 6-7 mm. Then the structure of the transition zones, general and local part of the structural components was examined.
3. Results and Discussion

Fig. 1 shows the morphology of the structures of BrZhNA 12-7-1 bronzes, manufactured in different ways: ingot crystallization with speed 15°C/min, casting with speed of a higher order [4], vacuum suction casting, in which the crystallization rate was not less than 700°C/s. In the cast the cross section of steel dendrites is 50-100 μm, and in this same alloy, obtained by melting and the vacuum suction, - more than an order of magnitude lower, but dendritic morphology and dispersion were substantially identical (Fig 1 b, c).

![Fig. 1 BrZhNA 12-7-1 bronzes structure: a) 3 kg cast ingot b) Ø = 7 mm casting rod obtained by vacuum suction c) argon arc melting similar bronze thickness 5 mm](image)

This suggests that the 7 mm diameter rod obtained by vacuum suction, experimentally models melted metal in its purest form. The fact is that the level of mechanical properties of melted metal obtained by conventional means, in a large extent depends on the technology, the conditions and the quality of deposition (defects, the effect of the substrate on which the casting is performed) and qualifications of the performer. The vacuum suction technology also allows to eliminate or significantly reduce the mentioned above negative factors and get the "idealized" melted metal, to which properties should be sought in practice. Mechanical properties of the investigated alloys are shown in Table 2 in comparison with bronze BrO10.

Table 2: Mechanical properties of experimental bronzes depending its manufacturing method.

<table>
<thead>
<tr>
<th>№</th>
<th>Bronze composition</th>
<th>Mechanical Properties</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>σ&lt;sub&gt;2&lt;/sub&gt;, MPa</td>
<td>σ&lt;sub&gt;0.2&lt;/sub&gt;, MPa</td>
</tr>
<tr>
<td>2-3 kg ingots, crystallization speed &lt; 15 °C/min</td>
<td>170</td>
<td>215</td>
<td>10-14</td>
</tr>
<tr>
<td>1</td>
<td>BrO10</td>
<td>220</td>
<td>295</td>
</tr>
<tr>
<td>2</td>
<td>BrZhNA13-6-1</td>
<td>346</td>
<td>336</td>
</tr>
<tr>
<td>3</td>
<td>BrZhNOA13-6-2-1</td>
<td>170</td>
<td>364</td>
</tr>
</tbody>
</table>

Note: Deviation of mechanical properties values from shown values is for ingots - 7% (items 2,3), for rods ≤ 3% (items 4,5)

Its friction coefficient is low in comparison with other bronzes, but it is not welded by melting methods, and has a low level of plasticity. Bronzes, containing about 2% tin (Table 2, items. 3, 5) have a reduced plasticity (especially low figures of relative elongation), but the level of resistance of these bronzes is higher than those of tin-free bronzes (the effect of solid-solution strengthening copper matrix with tin [10]). They have an unsatisfactory weldability and for electric arc welding methods are not applicable.

Thus, there is reason to consider bronze BrZhNA 12-7-1, obtained by vacuum suction - as "idealized" melted metal. Set of mechanical properties of bronze BrZhNA 12-7-1 (Table 2, item 4) is quite satisfactory: the level of its strength is higher than BrO10 bronze in 30-40%, and plasticity - in 3-4 times.

In [11] it was considered the formation during crystallization process of steel dendrites in bronzes, similar to those we are now considering, and it was shown the dendrite growth in the solid state as a result of the diffusion deposition Ni, Fe, Al from copper matrix to the surface of the dendrite. During casting of the composite bronze on BrZhNOA 12-7-2-1 (Fig. 2) takes place reverse process. Bronze matrix has a melting temperature of 1100°C and in the heat-affected zone at a distance up to 0.8 mm from fusion line in the molten metal steel dendrites partially dissolve (Fig. 2c).

![Fig. 2 Structure particularities of fusion zone in bronze BrZhNA 12-7-1 (cladding) on bronzes BrZhNOA 12-7-2-1 (a); dendrites morphology in the cast (b); dendrite surface dissolving in ingot (c) heat-affected zone.](image)

This is evidenced by the morphology and chemical analysis of the local zones 1-5 (Table. 3). Matrix, up to its fusion, had the composition of zone 2 (Fig. 2c), after dissolving in the molten matrix, steel dendrite surface and its subsequent crystallization, it is formed zone 3, enriched in Fe and Ni. Thus, the shell (zone 3), follow the contours of the dendrite (zone 4), whose morphology resembles eutectic and consists of a bronze matrix with the composition of the zones 1 and disperse steel inclusions measuring less than 2 μm (for comparison: the size of the primary matrix inclusions (zone 2) formed during the casting crystallization, not less than 5 μm). Such structure is characteristic of the entire heat-affected zone, subjected to melting during casting (Fig. 2a).

Table 3. Dendrite chemical composition in bronze BrZhNOA 12-7-2-1 and adjacent matrix, wt.%

<table>
<thead>
<tr>
<th>№ zone</th>
<th>Fe</th>
<th>Ni</th>
<th>Al</th>
<th>Sn</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>1.7</td>
<td>3.4</td>
<td>0.3</td>
<td>4.5</td>
<td>90.3</td>
</tr>
<tr>
<td>2</td>
<td>4.9</td>
<td>7.0</td>
<td>0.4</td>
<td>3.1</td>
<td>84.6</td>
</tr>
<tr>
<td>3</td>
<td>17.8</td>
<td>10.9</td>
<td>0.6</td>
<td>2.4</td>
<td>68.3</td>
</tr>
<tr>
<td>4</td>
<td>54.1</td>
<td>21.9</td>
<td>0.4</td>
<td>-</td>
<td>23.6</td>
</tr>
<tr>
<td>5</td>
<td>58.5</td>
<td>18.5</td>
<td>0.3</td>
<td>-</td>
<td>22.8</td>
</tr>
</tbody>
</table>

Next, we consider the possibility of casting the test composite BrZhNA 12-7-1 bronze on cast iron SCh28.

In Fig. 3 it is shown fusion zone of cast iron and bronze. The transition zone 1, formed during the crystallization process on iron base and differs a little in chemical composition of cast iron (Table. 4), however, in this region bronze inclusions can be observed (zone 2-1) and inclusions, close in composition to the dendrites (Fig. 1, e; Table 3, zone 5), but with a different morphology – undetermined (zone 2-3).

Bleached gray cast iron are not observed, what is evidenced by the low micro-hardness of the entire zone 1. Peripheral melted
bronze layer is closed, in chemical composition, to its original composition (zone 3-1), however, the inclusions of maraging steels have a globular form instead of a dendritic one (see Fig. 2b).

On the basis of available data, it should be assumed that such a change in the morphology of the steel inclusions in the melted bronze, as we suppose, is due to the presence in it of silicon (zone 3-2), which diffused form cast iron to the entire thickness of the bronze cladding. Formally, we can conclude that precisely silicon prevents the formation of dendrites.

The results show that: the friction coefficient of bronze BrZhNA 12-7-1 is slightly higher than BrO10 and the friction force values for welding and casting, obtained by vacuum suction, are the same (Figure 4, curves 1 and 2). Wear intensity of experimental bronze BrZhNA 12-7-1 (Table. 5, item 1, 2) is substantially (one order of magnitude) lower than that of similar castings (Table. 5, item 3). This is due to the high dispersion of the dendritic component (Figure 1, b, c) - the morphology of these states is identical and differs fundamentally from that of the casting (Figure 1, a). BrO10 bronze wear resistance is also significantly different from the cladding Table. 5, item 4).

These results once again confirmed that the method of casting manufacture by vacuum suction experimental models melted metal, at least for the group of studied experimental bronzes.

4. Conclusions

Our study showed that the experimental composite bronze BrZhNA 12-7-1 may be a prototype of industrial bronze, designed primarily for casting and welding products and assemblies operating under-sliding friction conditions. The lack of intermetallic compounds of group B and of tin in this bronze guarantees a higher level of mechanical, technological and service properties.

The method of experimental modelling of the melted deposited metal by vacuum suction proposed and justified in the paper enables to get production of castings, ideally corresponding to melted metal.

Preliminary studies have shown that at sliding speeds higher than 6 m/s it is possible to grasp this bronze to the counterbody. However, we have already tested the option of replacing the maraging steel type N16YU1 dendrites by stainless steel dendrites, suggesting that it can contribute to increase the permissible sliding speed.

5. References

11. Потехин Б.А., Илюшин В.В., Христолюбов А.С., Жиляков А.Ю. Формирование структуры композитных бронз, армированных стальными дендритами // Физика металлов и металлургия. 2014. Том 115. № 4. С. 442-448.

---

Table 4: Chemical composition of structural components in melted bronzes BrZhNA 12-7-1 on cast iron brand SCh28 (composition shown in table 4).

<table>
<thead>
<tr>
<th>№ zone</th>
<th>Fe</th>
<th>Ni</th>
<th>Al</th>
<th>Si</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>94.0</td>
<td>0.6</td>
<td>-</td>
<td>2.7</td>
<td>2.7</td>
</tr>
<tr>
<td>2-1</td>
<td>70.2</td>
<td>14.1</td>
<td>0.4</td>
<td>1.1</td>
<td>13.2</td>
</tr>
<tr>
<td>2-2</td>
<td>3.0</td>
<td>3.1</td>
<td>0.7</td>
<td>0.2</td>
<td>93.0</td>
</tr>
<tr>
<td>2-3</td>
<td>68.5</td>
<td>13.6</td>
<td>0.2</td>
<td>1.2</td>
<td>16.0</td>
</tr>
<tr>
<td>3-1</td>
<td>3.5</td>
<td>3.9</td>
<td>0.7</td>
<td>-</td>
<td>91.9</td>
</tr>
<tr>
<td>3-2</td>
<td>70.7</td>
<td>11.5</td>
<td>0.2</td>
<td>1.0</td>
<td>16.7</td>
</tr>
</tbody>
</table>

Thus, during casting on iron the composite bronze, reinforced with steel dendrites of maraging steel ceases to be composite, but conserves satisfactory technological properties.

The results of the friction coefficient measurements of bronze BrZhNA 12-7-1 obtained in many different ways, is shown in Fig. 4, and the wear resistance in Table 5.
INCREASING WEAR RESISTANCE OF MACHINES DETAILS FROM CARBON AND 
Cr – Mn – N STEELS OF THE COMPLEX BORIDE COATINGS

Prof. Dr. Chernega S.\(^1\) Poliakov I\(^2\), Krasovskyi M.\(^2\)

\(^1\)National Technical University of Ukraine "Kiev Polytechnic Institute", 03056, Ukraine, Kiev, Polytechnique st., Bldg. 9, e-mail: smchernega@mail.ru; polykov_igor@mail.ru

\(^2\)The Frantsevich Institute for Problems of Materials Science of NAS of Ukraine, 03680, Ukraine, Kiev, Krzhizhanovskogo st., 3

Abstract: The results of studies wear resistance of boron coatings alloyed with Cu, V, Nb, Cr or Ti under dry friction – slip in the air, and it was determined that the best coatings are the boride layers alloyed with copper. Alloying with copper increases wear resistance of boride layers in 3 times. When alloying of boride layers with niobium wear resistance increases in 2,8 times, with titanium – in 2,4 times, with chromium and vanadium – in 1,5 times. Wear resistance of boride coatings alloyed with copper under dry friction – slip 14 times higher than wear resistance of bronze of Cu – Al – Mn system and wear resistance increases when applying boride layers alloyed with copper.

KEYWORDS: BORIDING, BORON LAYER, COPPER, STRUCTURE, DIFFUSION, FRICTION, MICROSTRUCTURE, MICROHARDNESS, WEAR RESISTANCE.

1. Introduction

Modern production is makes high demands to increase the resource and reliability of machines and mechanisms. The improvement of these characteristics may subject the use of appropriate materials of construction, which would optimally combines high hardness, ductility, and hence the wear resistance. Therefore, one of the most important tasks facing materials scientists - improving the tribological characteristics of existing materials and coatings or creating new ones with the desired properties.

Surface saturation alloys simultaneously two or more elements (multicomponent diffusion saturation) has advantages as compared with the saturation of one element. It allows you to combine the properties of the resulting coating produced by individual elements with special properties of their compounds [1].

The multicomponent diffusion boride coatings have the whole complex of physical, chemical and mechanical properties, leading to high wear resistance, heat-resistant and corrosion-resistant products, subjected to chemical - thermal treatment. Multicomponent diffuse coating superior monocomponent coating by performance properties, with the technology of producing virtually no complicated [2-4].

The aim is to study the behavior of alloy boride coatings in conditions of contact wear at dry friction - slip, as well as the identification of the alloying elements affect on indicators wear resistance these coatings. It is known that as a result of saturation steels of boron can be formed surface layers with different structure and high complex properties. Therefore, improvement of surface alloying processes both Cr and B, B and Cu, B and Nb, B and V, B and Ti at the chemical-thermal treatment in the powder mixtures is an actual problem, which is theoretical and practical interest.

2. Materials and Experiment

Processes borating performed powder method in a special at a temperature of 975 °C during 4 hours using fusible valves. Saturation steels B and Cr, B and Cu, B and Nb, B and V, B and Ti performed in powder mixtures on the basis of technical boron carbide B_{2}C with the addition of powders Cu, V, Nb, Cr or Ti.

Microstructural and microhardness studies on the equipment PMT – 3 no less than 15 – 20 fields of view at a load of 0.49 – 0.98 N. Measuring accuracy microhardness was ±300 MPa.

Analyzed the phase composition of coatings for X-ray diffractometer DRON 2.0 in copper Kα1, Kα2 monochromatic radiation and determined their chemical composition analyzer «Camebax Sx50».

Testing of coatings on the wear resistance performed on friction machine. In a study as material of the body counter was used hardened and tempered steel U8, the test subject sample size 10x20x5mm of steel 45 after boriding with the addition of alloying metals in accordance chromium, vanadium, titanium, niobium, or copper.

3. Results and discussion

The diffusion saturation of carbon steel in boriding powder mixtures with the addition of powders Cu, V, Nb, Cr or Ti leads to the formation of boride phases FeB and Fe_{2}B thickness of 120 – 140 microns. Alloying elements formed corresponding substitutional solid solutions in the lattice of iron (Fe, Me)B, (Fe, Me)_{2}B. For example, the saturate the environment chromium powder leads to alloying of boride phase: (Fe, Cr)B, (Fe, Cr)_{2}B. At the same time there is a change of the crystal lattice periods boride phases. Alloying boride iron of chromium leads to a reduction volume of the rhombohedral cells dissolves the dissolution of chromium boride phase to 0.3 - 1.2 % at. The minimum volume the rhombic phase cells FeB obtained on the steel 45 corresponds 4 wt. %. Cr in the saturating environment is 65,763 · 10^{-3} nm\(^3\), whereas the volume of the initial phase is FeB 65,899 · 10^{-3} nm\(^3\). The main contribution in the decrease volume the unit cell rhombic makes a change parameter "c"). Comparison measurement results volume elementary rhombic lattices FeB phase and its microhardness depending on the chromium content in saturating environment showed an inverse relationship between these characteristics. The maximum value of the microhardness corresponds to the minimum value volume of the unit cell rhombic lattice FeB phase at 4 wt. % Cr in the powder environment at the saturation with boron and chromium on steel 45. Maximum values of microhardness diffusion layers are obtained at 4 – 6 wt. % Cr and up to FeB phase – 21,20 GPa, and for phase Fe_{2}B – 21,150 GPa on the steel 20, on steel 45 – 21.00 GPa and steel U8 – 18.4 GPa. Solar stratified X-ray phase analysis boride coatings allows the identification on the surface layer phase FeB, and beneath it Fe_{2}B phase. The most intense reflections from Fe_{2}B phase planes were reflexes (002) (112) (004) when taking the initial sample. Judging by the intensity indexes registered peaks phase Fe-B is located in textured state. Fixed decrease values periods rhombic lattice a, b, c FeB phase with increasing chromium content in saturating the environment. In this case the most intensively decreases periods "c" to the minimum value of 0.40575 nm FeB phase at 4 % Cr in the mixture, that corresponding the maximum deviation value period "c" (equal to 0.0005 nm) as compared with unalloyed FeB phase. Thus, the introduction of the alloying metal leads to changes parameters the crystal structure of boride phase and consequently changing microhardness, fracture toughness, distribution of residual stress, and other properties and characteristics of boride phases.

From the value of resistance plotted the graphics depending on the duration of wear and the distance traveled. The path is determined by knowing the number of turns in a given period of time. During one turn trip distance was 4 cm (because the length of the sample is 2 cm).

Kinetic curves wear boride coatings without alloying and alloyed accordingly V, Cr, Ti, or Cu, in conditions of dry slide friction are shown in Figure 1.
Thus, the wear rate of boride coatings alloyed with chromium in the first hour of wear was $2.08 \times 10^{-6}$ kg/m²·s, Ti – $1.18 \times 10^{-6}$ kg/m²·s, V – $1.88 \times 10^{-6}$ kg/m²·s, Nb – $5.56 \times 10^{-6}$ kg/m²·s, Cu – $0.972 \times 10^{-6}$ kg/m²·s, and without alloying $5.06 \times 10^{-6}$ kg/m²·s. These data show that the lowest wear rate to complete grinding period characterized boride coating obtained at complex saturation with B and Cu herewith the wear rate of iron boride saturated copper in the 2 times lower than saturated Cr and V and 5.2 times less than boride layers without saturation. Continue the wear up to 2 hours showed that the rate of wear of the surface layers during this period was: after boriding $3.13 \times 10^{-6}$ kg/m²·s, Ti – $0.763 \times 10^{-6}$ kg/m²·s, V – $1.88 \times 10^{-6}$ kg/m²·s, Cr – $1.98 \times 10^{-6}$ kg/m²·s, Cu – $0.659 \times 10^{-6}$ kg/m²·s. On step 5 hour the wear rate diffusion layers was accordingly after boriding $1.75 \times 10^{-6}$ kg/m²·s, Ti – $1.54 \times 10^{-6}$ kg/m²·s, V – $1.29 \times 10^{-6}$ kg/m²·s, Cr – $1.29 \times 10^{-6}$ kg/m²·s, Cu – $0.847 \times 10^{-6}$ kg/m²·s. At all stages of wear of the lowest wear rate detected in the boride layers obtained at complex saturation with B and Cu (Fig. 1). Increased of duration wear samples with boride layers shows the effect of saturated elements on the linear wear land. Studies have shown that the best wear resistance under dry friction - slip have boride phase obtained at complex saturation with B and Cu. It was determined that the coating with the lesser hardness of 14.5 GPa, saturated copper showed in 2 times better wear resistance compared with comparison investigated saturated elements and after boriding. This is because in the structure of boride layers detected separate copper inclusions (Fig. 2), which serve as the solid oil. If the test samples supply in a row at order to increase wear resistance, it will look like the following:

$\text{Fe}_2\text{B} \rightarrow (\text{Fe} + \text{V})_2\text{B} \rightarrow (\text{Fe} + \text{Cr})_2\text{B} \rightarrow (\text{Fe} + \text{Ti})_2\text{B} \rightarrow (\text{Fe} + \text{Nb})_2\text{B} \rightarrow (\text{Fe} + \text{Cu})_2\text{B}$

Comparison wear boride phases with bronze Cu – Al – Mn showed that the wear resistance of boride phase prevails over bronze in 30 times at complex saturation with B and Cu, without alloying – 15 times (fig. 5).

Comparison wear boride phase alloyed Cu, V, Nb, Cr, Ti and carbide coatings on the basis of titanium on steel 45 shown in fig. 6, fig 7 and 8.
In Fig. 6 shows the kinetic curves for wear boride and carbide coatings. Titanium carbide coatings on steel 45 without heat treatment had a hardness of 30 GPa, and a thickness of the diffusion layer 12 – 13 microns. Diffusion carbide coating on based titanium completely worn out after 20 minutes of testing. This is due to the small thickness of the diffusion layer and the low strength of the matrix, which was subjected to a heat treatment after deposition of diffusion carbide coatings. Thickness diffusion boride layers was 120 - 140 microns, titanium carbide - 12 – 13 microns. Diffusion carbide layers in a friction – slip sank into a soft plastic matrix during wear. Additional heat treatment leads to a redistribution of residual stresses in the diffusion layers, and carbide coatings might result to formation of cracks in the diffusion layer, so additional thermal treatment of carbide phases did not yield to.

In the paper shows that in the adopted severe conditions of dry friction - slip chemical heat treatment of boriding dominates over thin carbide coatings, which can be in 1.5 times higher microhardness, but an order of magnitude less thickness layer.
4. Conclusions

Investigated the wear resistance in the under the dry friction – slip on the air boride coatings saturated Cu, Ti, Cr, V, Nb and carbide coatings on the basis Ti on steel 45. It was established that the best coatings in the conditions of under dry friction - wear is boride phase FeB and Fe₂B obtained after complex saturation with B and Cu which increase the wear resistance without alloying boride layers in 3 times. Then, as carbide coatings in conditions of under dry friction - slip on air worn out after 20 minutes of research. Wear resistance of boride layer in 30 times higher than the wear resistance of bronze.

5. References