WEAR PROPERTIES OF Al₂O₃/SiCp COMPOSITES

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Keywords: Al₂O₃/SiC, tribology, wear, ball-on-disc

1. Introduction

The properties of advanced ceramics that make them suitable for tribological applications are low density, high hardness, compressive strength and corrosion resistance over a wide temperature range. Alumina based ceramics are among the last years increasingly applied as wear parts in different applications. The wear properties of pure alumina and ceramic nanocomposites are determined by numerous characteristics, for example grain size, modes of microcracks propagation, grain dislodgement and pull-out, weight loss, wear rate, residual porosity and stresses.

Alumina/SiC nanocomposites are much more resistant to severe wear than monolithic alumina. The wear properties depend strongly on the grain size in pure alumina, but in the case of nanocomposites, these are reported to be independent on the size of alumina grain. A. Limpichaipanit and R. I. Todd tested the nanocomposites Al₂O₃+5 and 10 vol.% SiC in abrasive wear and observed improvement in wear resistance which stems from a reduction in the size of the individual pullouts caused by brittle fracture owing to the accompanying change in fracture mode. They reported that a large fraction of the worn surface of alumina has been removed by “pullout”. The nanocomposites display smoother worn surfaces with much less pullout.

The aim of this investigation is to study the influence of SiC addition on the wear behaviour of Al₂O₃/SiC nanocomposites at room and high temperatures up to 700 °C.

2. Experimental procedure

The SiC reinforced Al₂O₃ composites used for the present study were prepared by hot-pressing of a pre-pressed powder mixture in a graphite mould at a temperature of 1550 °C for 5 min and a pressure of 30 MPa in a static argon atmosphere with overpressure of 15 kPa. In all hot pressing runs, a pressure was applied at 900 °C and was maintained during heating. The heating and cooling rates were 30 and 20 °C min⁻¹, respectively. The pressure was released during cooling at 1200 °C and specimens were left to cool in the die to room temperature. The resulting composite materials contained 5 % (AM5), 10 % (AM10), and 20 % (AM20) volume of SiC phase.

Tab. I illustrates some mechanical properties of the materials used in this investigation. The wear behavior of the materials was studied by unlubricated ball-on-disk experiments (on the high-temperature tribometer DTHT 70010, CSM Instruments, Switzerland), as shown in Fig. 1. The conditions of the experiment: applied load 5 N, sliding distance 500 m, sliding speed 0.1 m s⁻¹ and temperatures 25 °C, 300 °C and 700 °C. The friction coefficients were continually recorded during the tests and wear volume on each specimen was calculated from the surface profile traces (at least 4) across the wear track and perpendicular to the sliding direction using the profilometer (Mitutoyo SJ-201, USA). The wear tracks were then examined by SEM to investigate the wear mechanisms.

The specific wear rate (r) is given by:

\[
 r = \frac{V}{FL} \quad \text{(1)}
\]

where \( F \) is the loading force [N]; \( L \) is total sliding distance [m]; and \( V \) is worn volume [mm³].

\[
 V = A(2\pi r)^2 \quad \text{(2)}
\]

where \( A \) is the average value of four different measurements of cross section area of the wear track estimated by profilometer [mm²]; and \( r \) is the sliding radius [mm].

Table I

<table>
<thead>
<tr>
<th>Samples</th>
<th>Density [g/cm³]</th>
<th>Hardness [GPa]</th>
<th>HV1</th>
<th>HV5</th>
<th>KIC [MPa m¹/²]</th>
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</thead>
<tbody>
<tr>
<td>AM</td>
<td>4.008</td>
<td>19.2 ±0.9</td>
<td>17.1 ±0.6</td>
<td>3.44 ±0.2</td>
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<tr>
<td>AM5</td>
<td>3.947</td>
<td>19.8 ±0.6</td>
<td>17.4 ±0.7</td>
<td>3.25 ±0.2</td>
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<tr>
<td>AM10</td>
<td>3.884</td>
<td>21.9 ±0.4</td>
<td>18.7 ±0.5</td>
<td>3.04 ±0.1</td>
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<tr>
<td>AM20</td>
<td>3.748</td>
<td>21.2 ±0.4</td>
<td>18.2 ±0.3</td>
<td>3.45 ±0.1</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 1. Schematic illustration of ball-on-disc type wear test
3. Results and discussion

In Fig. 2 characteristic microstructures of the monolithic alumina and the Al₂O₃ + 5 vol.% SiC composite is illustrated. The alumina consists of grains with size of several micrometers with maximum size up to 10 µm. The composite exhibits significantly finer microstructure in comparison to the alumina with the size of grains approximately 5 µm. Silicon carbide nanoparticles are visible in the microstructure of the composite, located inter and intragranularly, with the size from several tens to several hundreds of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers. No cluster of SiC nanoparticles has been found in the microstructure of composites with low volume fraction of micrometers.

Fig. 3 illustrates the temperature dependence of friction coefficient during the sliding distance of 500 m. As it is visible the friction coefficient increased with increasing temperature. At room temperature the friction coefficient of the composites is similar, and exhibits a value around 0.35. The coefficient of friction of alumina is higher, about 0.4. The friction coefficients at 300 °C and at 700 °C are significantly higher for all materials. The tendency valid at room temperature seems to be retained, the friction coefficient decreases with increased SiC content. The lowest friction coefficient at 700 °C was found for the system of Al₂O₃ + 20 vol.% SiC with the value of 0.65.

During the tribology tests of the composites at room temperatures and 300 °C the wear volume was very low and the wear tracks couldn’t be measured by the profilometer because of their small size. In the monolithic alumina the specific wear rate at the temperature 300 °C was 38.4x10⁻⁵ mm³ N⁻¹ m⁻¹. In all Al₂O₃/SiC composites the wear resistance seems to be significantly higher in comparison to the monolithic alumina.

Comparison of the wear rates at 700 °C is illustrated in Fig. 4. The beneficial effect of the higher volume fractions of SiC particles is clearly visible. In spite of the fact that there is no significant difference in the hardness and fracture toughness values of the monolithic alumina and Al₂O₃ + 20 vol.% SiC composite the difference in wear resistance is significant. The wear resistance in AM10 and AM20 composites is approximately three times higher than that of the monolithic alumina.

Our results are in a good agreement with the results of Guicciardi et al.⁵ who investigated the influence of the microstructure on wear behavior of Al₂O₃-based composites, (with...
5 vol.% SiC sub-micron-sized particles) tested at RT, 40N-0.5 m s\(^{-1}\), 10 km. They reported the values of friction coefficients from 0.6 to 0.8 and disk specific wear from \(0.9 \sim 2.0 \times 10^{-6}\) g km\(^{-1}\) N\(^{-1}\) and pin specific wear \(1 \sim 3\) g km\(^{-1}\) N\(^{-1}\). In the case of pure alumina the test wasn’t finished because of high values of friction coefficient (about 1.0) due to the rough surface of wear track. In the limited portions recorded, the disk/pin wear for this distance was about \(1050\) g km\(^{-1}\) N\(^{-1}\) and \(75\) g km\(^{-1}\) N\(^{-1}\), respectively.

Examples of the worn surfaces of the monolithic alumina and \(\text{Al}_2\text{O}_3 + 5\) vol.% SiC composite tested at room temperature are given in Fig. 5. The difference in the damage mechanisms is visible. The worn surface of the composite is smoother, little damaged with less microcracking and fewer scale-like features typical for fatigue processes in cyclical wear. The SiC particles seem to enable easier particle pull-out and self polishing of the worn surface. Fig. 6 and 7 illustrate a comparison between the wear tracks of \(\text{Al}_2\text{O}_3/\text{SiC}\) composite with 20 vol.% of SiC particles after the test at elevated temperatures. The tracks produced at 300 °C are very faint and shallow with periodic microcracking perpendicularly located to the direction of main track. The distance between the microcracks is approximately 15 μm. The damage in the material between microcracks is significantly lower. The main wear mechanisms identified were grain fracture and grain comminution at the temperature 700 °C. The wear track of alumina produced at 700 °C was wider and more significant than that in the composite with 20 vol.% SiC (see Fig. 7).

### 4. Conclusions

The influence of SiC addition on the wear behaviour of \(\text{Al}_2\text{O}_3/\text{SiC}\) nanocomposites at room and high temperatures up to 700 °C has been investigated. The wear resistance of the monolithic alumina was significantly lower than that of the composites at room and elevated temperatures. Plastic deformation, combined with brittle fracture was identified as the main wear mechanisms. Their intensity was different and correlated to the wear resistance of the materials.

The coefficient of friction of the composites was lower in comparison to that of monolithic alumina at all temperatures. At the high temperatures during the friction process, microcracks develop due to surface fatigue, which are the origin of severe wear damage during the test.

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### REFERENCES

E. Csehová*, J. Dusza¹, P. Hvizdoš⁵, J. Sedláček⁷
¹ IMR, SAS, Košice, Slovak Republic, ⁵ Institute of Inorganic Chemistry, SAS, Bratislava, Slovak Republic: Wear Properties of Al₂O₃/SiC⁴⁰ Composites

The influence of SiC addition on the wear behaviour of Al₂O₃/SiC nanocomposites at room and high temperatures up to 700 °C has been investigated. The presence of SiC particles has a positive effect on the room and high temperature coefficient of friction of Al₂O₃/SiC nanocomposites. The wear resistance of composites was significantly higher at all temperatures in comparison to the monolithic alumina. Plastic deformation, combined with crack formation and brittle fracture were the main wear mechanisms.
MICROSTRUCTURE AND PHASE ANALYSIS OF 0.5Cr-0.5Mo-0.25V CREEP-RESISTANT STEELS AFTER LONG-TERM SERVICE

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Key words: Low-alloy steels, microstructure, phase analysis, thermodynamic calculations

1. Introduction

Low alloy 0.5Cr-0.5Mo-0.25V steels have been extensively used in power industry in the form of pipes and forgings approximately from the 60-ties in the last century. In spite of their relative mild alloying they are able to operate in conditions of the high temperatures, up to 580 °C, and increased pressures. This is due to the long term microstructure stability during service that is achieved by proper heat treatment (HT), the normalizing at 950–990 °C and tempering between 600–730 °C. The chosen HT procedure results in a complex mixed microstructure (ferrite + pearlite, ferrite + bainite, bainite, bainite+martensite) with primary and/or secondary particles of the cementite, molybdenum-rich carbides M2C and M6C, (Cr, Fe)-rich carbides M23C6 and M7C3 and complex MX carbonitrides. During the exploitation at high temperatures the particles can precipitate, grow, and dissolve in dependence on their thermodynamic stability. Moreover the particles simultaneously interact with dislocations what is one of the crucial phenomena influencing the creep behaviour of the material. Precipitate characteristics like the size, morphology, distribution, chemical composition etc. are the variables used in correlation with mechanical properties and their degradation.

Also for the reliable component life design and the precise remnant life estimation of the component, one should know what exactly happens with individual phases in the respective alloy and environment conditions. This kind of the material is used also in atom power stations where the detail microstructure analyses after the exploitation are done with the intention to prolong a component life-time via thermo-mechanical treatment, e.g. (ref⁴). The present study is a part of the work dealing with the phase analysis in ex-service materials for energy industry after 17–20 years in service.

2. Experimental

For experiments the samples of six steels aged for almost 200,000 h at 568 °C in service (JP at 540 °C) were used.

Chemical composition is given in Table I. All samples were chosen from the areas of the parent metal apart the weld joints.

Microstructure of the materials was investigated by LM and analytical STEM. The EDXS and SAD methods were used for the identification of particles in carbon replicas. For an equilibrium calculation of the phases and their composition in the materials a software Thermo-Calc was employed.

3. Results

3.1. Hardness and microstructure

Materials 12, 14, 15 achieved the HV10 values of 130, for samples 16, 24, and JP they were higher, 136, 148, and 179, respectively. The microstructure of main steam pipes materials 12, 14, 15, 16 is tempered, consisting of ferrite (F) and particles of two size categories (see Fig. 1). The 1–10 µm large particles are embedded at the grain boundaries, while there are smaller particles in the matrix. In vicinity of the large particles, the precipitate-free areas can be seen. The character of sample 24 from reheat pipe-work is similar, only the particles on the grain boundaries are smaller, up to app. 5 µm.

A banded microstructure has been revealed in the sample JP, Fig. 2. The wide bands of ferrite alternate with narrow bands of spheroidized pearlite (P).

![Illustration of the tempered microstructure of materials 12-16 consisted of ferrite-carbide mixture, SEM](image-url)
3.2. Phase analysis

Large particles at the grain boundaries have been analyzed as (Fe, Cr, Mo)-rich particles of the M$_2$C$_6$ type. This kind of particles is present in all materials except for 15 one.

Densely precipitated particles in the matrix are of two types, V-Mo-rich MX carbides and/or carbonitrides and Mo-V rich particles of M$_2$C$_6$, Fig. 3a, c, d. They were found arranged in the form of so called H-type carbides, notable in the samples 12–16V, Fig. 3b.

Moreover the fiber-shaped VC has been found, Fig. 4b. JP sample differs from the others by the presence of coagulated cementite in the localities of spheroidized pearlite, Fig. 4a.

The calculated phase diagram for the systems corresponding to the experimental steels is shown in Fig. 5a, except for the steel JP the phase diagram of which is in Fig. 5b.

Fig. 2. Metallography of tempered banded ferrite+pearlite microstructure in sample JP, a – LM, b – SEM

Fig. 3. Fine precipitation in the matrix. a – overall view, b – H-carbides, c – VC with SAD, f-rod-like M$_2$X with SAD

Fig. 4. a – Cementite in JP sample with SAD TEM, carbon replica, b – Fibrous form of VC carbide

Fig. 5. Calculated diagrams for a – steel 15, b – steel JP

4. Discussion

The presence of spheroidized pearlite in the microstructure of pipes suggests that a slow cooling rate was used at their production. Any of the investigated steels does not contain cementite in equilibrium conditions. The steel should contain less Cr to get close to/into the area representing a system with cementite as the equilibrium phase, Fig. 5b. This example supports the idea, that a relatively small change in the chemical composition together with the HT used may have a decisive influence on the microstructure of the low alloy steels.

There is a large variety of (V,Mo)(C,N) particles in the investigated materials. A part of MX carbonitrides occurs in the form of larger particles (100–300 nm) compared to the smaller ones (30 nm). This kind of bimodal MC particles distribution is described by Senior. The larger square shaped particles often form so called H-carbides, when Mo$_2$C rods nucleate at the sides of MX particles enriched with Mo from the solid solution, and grow partly at their expense. According to Carruthers and Collins, VC is intensively exchanged by Mo$_2$C up to 600 °C. In (ref. 7) Collins shows that Mo$_2$C is major carbide in a similar material after 100,000 h aging at 570 °C. The phenomenon is explained in (ref. 7) by the change of thermodynamic stability of the mentioned carbides at temperatures around 600 °C. MC carbide is the first carbide to precipitate thanks to the chemical activity of V and to the fact that its formation minimizes the chemical potential of C (ref. 7). It can precipitate as the inter-phase precipitate during γ→α transformation, randomly dispersed precipitates, on dislocations in remaining austenite, and occasionally in fibrous form. The inter-phase and fibrous forms are known to deteriorate the toughness of the materials. Small areas of the fibrous MC were found in our JP and 16 samples, Fig. 5.

Chemical composition of MX and M$_2$C particles is plotted in pseudoternary phase diagram, Fig. 6. All data points lie...
around a tie line joining the V and Mo corners and document the variability of V/Mo ratio in both particle types, 1.0÷6.3 in MX and 0.2÷0.5 in Mo 2C before achieving the equilibrium state. The ratio calculated for equilibrium conditions is app. 1.6 for MX and 0.05 for Mo 2C. In both cases a very small amount of Nb was detected.

M23C6 carbides precipitate at the grains of former austenite and during annealing are growing to large sizes, Fig. 1. In all experimental steels the carbide contains 73–84 wt.% of Fe in the metallic part. These values are roughly comparable with the calculated equilibrium composition (app. 65 wt.%).

The experimental and calculated equilibrium phases at the exploitation temperatures are compared in Table II. The presence of Cr-rich particles in both forms is common for the studied steels and depends on the chemical composition, Fig. 6. Moreover as mentioned in [12], the M-Cr carbide precedes the M2C6 one in low alloy steels, and the higher Cr/Mo ratio supports the existence of M2C6. The composition very close to the line between two phase areas can explain the presence of cementite in the JP material, Fig. 5b.

Concerning the hardness, the relatively small values after the service could be explained by the reduction of solid solution strengthening. This results from Mo depletion from solid solution due to the precipitation of Mo2C carbide particles. The highest hardness of JP material can be related to the shorter time and lower temperature of service and hence more solute remaining in solid solution.

5. Conclusions

The detailed study of phases precipitated after long term exploitation can be summarized as follows:

The 1/2Cr-1/2Mo-1/4 V alloys consist in equilibrium of ferrite, MX, Cr-rich M7C3 or M23C6 and Mo2C phases. The chemical composition of carbides after the exploitation at high temperatures is comparable with that calculated for equilibrium conditions.

Bimodal distribution of MC particles was observed. MX varies from V-rich to V,Mo-rich carbide and continuously changes to M2C with variable Mo-V content.

The work has been carried out partly in the SFP project SmartWeld, VEGA project No. 2-0128-10 and APVV SK-SL-0029-08.

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A. Výrostková, M. Svoboda, V. Homolová, L. Faliat, J. Kepič (*Institute of materials research, Slovak Academy of Sciences, Košice, Slovakia, 1 IPM ASCR, Brno, ČR): Microstructure and Phase Analysis of 0.5Cr-0.5Mo-0.25V Creep-Resistant Steels After Long-Term Service

The work characterizes the microstructure and phase analysis of the precipitates present in six low-alloy 0.5Cr-0.5Mo-0.25V creep-resistant steels after long-term service. Four samples were from main steam piping and two samples from re-heat pipe-work. Chemical analysis and identification of the precipitates were carried out by energy dispersive X-ray spectroscopy (EDXS) and selected area electron diffraction (SED), respectively. Thermodynamic calculations for the studied 0.5Cr-0.5Mo-0.25V steels were performed by software Thermo-Calc. The steels in equilibrium state consist of ferrite, MX, Cr-rich M7C3 or M23C6, and Mo2C phases.

Table II

<table>
<thead>
<tr>
<th>Sample</th>
<th>Calculated</th>
<th>Measured</th>
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<tr>
<td>12</td>
<td>F+MX+2+23</td>
<td>F+MX+2+23</td>
</tr>
<tr>
<td>14</td>
<td>F+MX+2+7</td>
<td>F+MX+2+23</td>
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<tr>
<td>15</td>
<td>F+MX+2+2</td>
<td>F+MX+2</td>
</tr>
<tr>
<td>16</td>
<td>F+MX+2+7</td>
<td>F+MX+2+23</td>
</tr>
<tr>
<td>24</td>
<td>F+MX+2+23</td>
<td>F+MX+2+23</td>
</tr>
<tr>
<td>JP</td>
<td>F+MX+2+7</td>
<td>F+MX+2+23</td>
</tr>
</tbody>
</table>

F – ferrite, 2 – M4C, 3 – M6C, 7 – M23C6, 23 – M23C6

Fig. 6. MX and M2C carbide composition changes
VPLYV TRYSKANIA NA ELEKTROCHEMICKÉ CHARAKTERISTIKY POVRCHU HORČÍKOVEJ ZLIATINY AE21

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Kľúčové slová: horčíková zliatina, elektrochemické charakteristiky, tryskanie, elektrochemická impedancne spektroskopia

1. Úvod


Legovým zlatím na báze Mg-Al kovových zlútiin sú významné v konštrukčnej teoretike, tryskanie, elektrochemická impedancne spektroskopia (EIS), ktoré boli realizované v Laboratóriu korózie Katedry materiálového inžinierstva SjF ŽU v Žiline na koróznom meracom systéme Voltalab 10. Brusené a tryskané povrchy boli pred EIS meraním exponované v 0,1M NaCl od 5 min do 168 hodín. Samotné EIS meranie prebiehalo vo frekvenčnom rozsahu 100 kHz – 50 mHz so zmenou 20krát na dekádu. Amplitúda striedavého napätia bola 20 mV, teplota pri otáčkovom pomerí 22 °C a meraná plocha 1 cm2. Výsledkom EIS merania bolo určenie elektródneho zloženia a elektródneho zloženia ktoré bolo použité pomocou programu EC-Lab (Bio-Logic SAS France).

2. Experimentálny materiál

Ako experimentálny materiál sme použili horčíkovú zlatinu AE21 (zloženie podľa ASTM: 2 hm.% Al, 1 hm.% RE, 0,3 hm.% Mn, zvyšok Mg) po priamnom extrudovaní, ktoré bolo realizované v Laboratóriu Technickej univerzity v Clausthale, Nemecko, pri 370 °C s extrúznym pomerom 22. Tryskanie povrchov experimentálnych materiálov bolo uskutočnené na Katedre technológií a materiálov SjF TU Košice na pneumatickom tryskacom stroji TJVP-320, pri tlaku 0,4 MPa a vzdialenosti dôzy od povrchu vzorky 200 mm. Ako tryskací prostriedok bola použitá sôda (NaHCO3).

3. Experimentálne metódy

Na stanovenie elektrochemických charakteristík povrchov AE zlatín bolo použitá metóda elektrochemického impedancne spektroskopie (EIS). Merania boli uskutočnené v Laboratóriu korózie kovov Katedry materiálového inžinierstva SjF ŽU v Žiline na koróznom meracom systéme Voltalab 10. Brusené a tryskané povrchy boli pred EIS meraním exponované v 0,1M NaCl od 5 min do 168 hodín. Samotné EIS meranie prebiehalo vo frekvenčnom rozsahu 100 kHz – 50 mHz so zmenou 20krát na dekádu. Amplitúda striedavého napätia bola 20 mV, teplota pri otáčkovom pomerí 22 °C a meraná plocha 1 cm2. Výsledkom EIS merania bolo určenie elektródneho zloženia ktoré bolo použité pomocou programu EC-Lab (Bio-Logic SAS France).

4. Výsledky experimentov

Výsledkom EIS meraní boli tvorená polodruhých znám v frekvenčnom rozsahu 100 kHz – 50 mHz so zmenou 20krát na dekádu. Amplitúda striedavého napätia bola 20 mV, teplota pri otáčkovom pomerí 22 °C a meraná plocha 1 cm2. Výsledkom EIS merania bolo určenie elektródneho zloženia ktoré bolo použité pomocou programu EC-Lab (Bio-Logic SAS France).

Pri analýze boli použité ekvivalentné obvody na obr. 4, ktoré boli definované na základe experimentalnych prechádzok. Ø je odpor elektrolytu, R1,2 sú polarizačne odporové časti impedancie systému (Bode magnitude plot) a fázového posunu (Bode phase plot) od frekvencie. Namerané Bodeho diagramy boli analyzované pomocou programu EC-Lab (Bio-Logic SAS France).

Obr. 1. Mikrostruktúra Mg zlatiny AE21, lept. kyselina pikrová, svetelná mikroskopia
Tabuľka I
Polarizačné odpory [Ω cm²] brúsených a sódou tryskaných povrchov v prostredí 0,1M NaCl po rôznych časoch expozície

<table>
<thead>
<tr>
<th>Čas expozície</th>
<th>Brúsený povrch</th>
<th>Sódou tryskaný povrch</th>
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<tbody>
<tr>
<td></td>
<td>Rp1 [Ω cm²]</td>
<td>Rp2 [Ω cm²]</td>
</tr>
<tr>
<td>5 min</td>
<td>124±18</td>
<td>50±5</td>
</tr>
<tr>
<td>1 h</td>
<td>164±5</td>
<td>42±8</td>
</tr>
<tr>
<td>4 h</td>
<td>273±3</td>
<td>143±30</td>
</tr>
<tr>
<td>8 h</td>
<td>361±21</td>
<td>250±4</td>
</tr>
<tr>
<td>16 h</td>
<td>571±53</td>
<td>-</td>
</tr>
<tr>
<td>24 h</td>
<td>492±17</td>
<td>-</td>
</tr>
<tr>
<td>48 h</td>
<td>169±36</td>
<td>-</td>
</tr>
<tr>
<td>96 h</td>
<td>99±2</td>
<td>19±2</td>
</tr>
<tr>
<td>168 h</td>
<td>120±19</td>
<td>23±5</td>
</tr>
</tbody>
</table>

Obr. 4. Ekvivalentné obvody použité pri analýze nameraných EIS kriviek
Dátum namerané na brúsených vzorkoch po 16, 24 a 48 hodinách boli analyzované použitím obvodu na obr. 4a. Ostatné krivky získané na brúsených vzorkoch a všetky krivky z tryskaných vzoriek boli analyzované pomocou obvodu na obr. 4b. Najdôležitejšou elektrochemickou charakteristikou, ktorú je možné stanoviť na základe analýzy Bodeho diagramov, je celkový polarizačný odpor (RP), ktorého hodnotu je rovná súčtu čiastkových polarizačných odporov Rp1 a Rp2. Hodnoty výsledných polarizačných odporov korózných systémov po rôznych časoch expozície sú v Tab. I.

5. Diskusia

Z analýzy Bodeho diagramov je zrejmé, že sa polarizačný odpor brúsených povrchov Mg zlatina AE21 zvyšuje s rastom doby expozície do 8 hodín. Odpor Rp brúsených povrchov je 3,5násobne vyšší po 8 hodinách expozície oproti predobrúseným povrchoch. Pri dlhších dobach expozície dochádza k zmene tvare kriev vymenchaných diagramov, pričom odpor Rp klesá. Je to spôsobené odpadávaním slabšej priťažlivých korózných produktov, ktoré majú čiastčne ochraný charakter, čím sa obnažuje aktívny povrch materiálu, dohľadá k jeho priamemu kontaktu s agresívnym prostredím a aktivnemu rozvoji korózného procesu. Tento odpor je rovný zmiešanému odporu prenosu náboja a vrstvy korózných produktov, ktoré sa v štych fázach korózného procesu nedajú na nameraných diagramoch jasne odlišiť. Ďalším rozvojom korózne dochádza k lokalizácii korózného napadnutia povrchu a zvýši sa, výsledkom zväčšania odporov na povrchoch. Priľahlá podmienka rozdelenia námecovych, koróznych, a koróznich produktov po porušení vrstvy; ušiel, že v prípade brúsených povrchov dochádza k ich ceľkovjej degradácii a strate schopnosti odolávať korózií v 0,1M NaCl. Otryskaný povrch si túto schopnosť zachováva až do 168 hod. expozície.

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LITERATÚRA

Evaluation of the electrochemical characteristics of AE21 magnesium alloy is presented in the article. The surfaces of tested alloys were treated by grinding and grinding followed by sodium bicarbonate shot-peening. The specimens were evaluated in 0.1 NaCl by electrochemical impedance spectroscopy method in combination with exposure tests. The exposition times were from 5 minutes to 168 hours. The measured data were analysed using equivalent circuits and polarization resistances of the various corrosion systems were determined.