Effect of Milling Conditions on the Microstructural Design in Aluminum Based Alloy Fabricated by SPS

Anton Smirnov 1,*; Ekaterina Kuznetsova 1; Yuri Pristinskiy 1; Pavel Podrabinnik 1,*; Alexander Mironov 1,2; Iosif Gershman 1,2; and Pavel Peretyagin 1,*

1 Spark Plasma Sintering Research Laboratory, Moscow State University of Technology “STANKIN”
Vadkovsky per. 1, Moscow 127055, Russia; e.kuznetsova@stankin.ru (E.K.); y.pristinskiy@stankin.ru (Y.P.);
p.podrabinnik@stankin.ru (P.P.); a.mironov@gmail.com (A.M.); isgershman@gmail.com (I.G.);
p.peretyagin@stankin.ru (P.P.)
2 Department of Scientific Research Programs, Grants and Projects, Railway Research Institute JSC “VNIIZHT”, 3rd Mytishchinskaya Street 10, Moscow 107996, Russia
* Correspondence: a.smirnov@stankin.ru; Tel: +7-4999-7323-70

Received: 8 October 2019; Accepted: 26 October 2019; Published: 29 October 2019

Abstract: In this work, aluminum-based alloys were mechanically alloyed by various doping elements followed by spark plasma sintering. Three different mixing methods were studied. First, all elemental powders were mixed for 24 h. In the second and the third method, powders were divided into three groups, and each group was milled separately for different times. Then the obtained mixtures were grouped together and again milled for 12 and 24 h, for the M2 and M3 mixtures, respectively. The influence of milling parameters on the microstructural features of sintered samples was determined using a scanning electron microscope (SEM) equipped with an energy-dispersive X-ray spectroscopy system. The correlation observed between the milling time and microstructural evolution of the powder particles during the milling process was to determine the optimal process parameters. For the first time, the present research has demonstrated the effectiveness of the milling process for the preparation of tailored mixtures of metallic powders, enabling microstructural characteristics that could be favorable for the formation of secondary structures on the tribosurfaces.

Keywords: Al-Sn-Pb alloy; mechanical alloying; spark plasma sintering; SEM-EDX; microstructure

1. Introduction

Aluminum (Al)-based alloys are widely used in engine building as sliding bearing materials for supporting the reciprocating motion of the crankshaft within internal combustion engines, owing to its excellent friction and wear properties, particularly when these alloys alloyed with a soft phase such as lead (Pb) and tin (Sn). A great effort is made to promote the properties of Al-based bearing alloys to satisfy the application requirement towards higher load and speed, together with eliminating toxic Pb, due to environmental concerns [1,2]. However, tin is a costlier and scarcer material, and it cannot replace lead, because every element has a favorable effect on the formation of secondary structures on the rubbing surface; therefore lead is still used for Al-based bearing alloys.

Thus, the main challenges for improving the properties of these alloys consist of two main aspects:

1. The strength of Al-based bearing alloys is generally low and cannot satisfy the high load requirement. It is, therefore, necessary to improve the material properties by adding some alloying elements. However, the presence of these elements can cause intensive wear of the steel counterpart. Besides, these alloys cannot be used as a substitute for bronze monometallic bearings.

2. Therefore, to form tribofilm, which carries outwear protective functions on the rubbing surface, the content of elements that allow the presence of soft components in Al-based bearing alloys should be increased.
The microstructure modifications and enhanced properties in Al-based alloys can be achieved by adding the elements Zn, Mg, Cu, Si, Mo, Ni, Cr and Mn. However, their concentration should be selected, based on not only their strength value, but also antifriction properties [3].

Additionally, each doping element has their unique properties, and consequently, their synergy should be taken into account.

For example, it was found that the poor wetting and strength of Al alloys can be improved by adding Mg to disrupt the passivating oxide layer of the Al particles [4]. Besides, the addition of Mg leads to the formation of a milled Mg–oxide layer on the worn surface of Al alloys, and can lower the junction strength, the adhesion with steel counter-face and this therefore decreases the friction coefficient effectively [5].

The addition of silicon to aluminum reduces melting temperature and improves fluidity. Silicon in combination with magnesium produces a precipitation hardening heat-treatable alloy. Moreover, silicon improves strength and resistance to abrasive wear as well.

Copper is one of the primary alloying elements for aluminum, based largely on the substantial age-hardening response of Al–Cu alloys. Moreover, copper has also been usually added to bearing alloys with a view to strengthening the aluminum matrix, thus improving mechanical properties and leading to a microstructural combination of a strong matrix with dispersed areas of a soft Sn phase [6–10]. The strengthening is maximum between 4 and 6% Cu, depending upon the influence of other constituents present.

The addition of zinc to aluminum (in conjunction with some other elements, primarily magnesium and/or copper) produces heat-treatable aluminum alloys of the highest strength. The zinc substantially increases strength and permits precipitation hardening. The low melting zinc phases show chemical reactivity that contributes to the formation of protective secondary structures on the rubbing surface, and increases conformability and wear resistance.

Introducing of molybdenum affects the microstructure and improves the strength characteristics of the aluminum matrix [11,12]. Moreover, Mo with its ability to form MoS₂ on the friction surface under lubricated conditions is a favorable additive in obtaining secondary structures. Besides, a previous study has shown that the presence of Mo and Cr acted as a favorable effect on the wear resistance of steel and the friction coefficient of the rubbing surfaces due to the formation of secondary structures with optimal composition [13]. However, the question about the desirability of using these metals in Al-based alloys is multifaceted and complex, and will be considered following a further detailed investigation of the mechanical and tribological behavior of obtained alloys.

In that way, the properties of Al-based alloys can vary substantially, depending primarily on the presence and concentration of alloying elements. However, it is important to take into account, not only the chemical composition of these alloys, but also the distribution, size and morphology of different phases. Moreover, aluminum matrix, solid inclusions and soft phases should be considered separately.

It is well known that the processing has a significant influence on microstructure and mechanical properties, and it is important to adopt a technique where a uniform dispersion of all constituents and good interfacial bonding is obtained.

Currently, the foundry is the most used and cost-effective method of production of Al-based alloys. However, this technique does not allow for variations of the content of different chemical elements and obtaining phases other than phase diagram, and it is difficult to account for the waste of low-melting elements (Mg, Pb, Zn, Sn).

While the powder metallurgy method does not have these disadvantages, mechanical alloying technology is an effective way to create forced miscibility in immiscible systems with controllable microstructure, which cannot be obtained by other means. Through this technology is it possible to affect the different phases that form in the alloy by changing the compositions of the elemental powders and their milling technology.

In the next step to prepare bulk alloy from the obtained mixtures, pressure-assisted sintering methods can be used. Spark Plasma Sintering (SPS) is a promising consolidation technique to
produce materials with a high density. The SPS technique works largely on the mechanism of electric spark discharge, where high energy pulsed current briefly generates and discharges spark plasma between the powder particles, resulting in high localized temperatures. The heat generated at the point contacts of the powder particles presents a surface treatment effect on the particle surfaces, removing contaminants and breaking the characteristic oxide layer on the aluminum powders before local welding or neck formation. This results in cleaner particle surfaces, an increase in the densification of the powders and an obvious comparative improvement in the properties of the resulting composites [14–18]. Additionally, the high heating rate—up to 1000 °C/min—allows significantly improved densification kinetics with lower time and temperature compared to those required for conventional sintering processes.

Multicomponent cast aluminum alloys with enhanced mechanical and tribological properties, as compared to Al-Sn and bronze alloys, were recently developed as candidates for monometallic journal bearings in the engine [19–23]. Based on these data, Al-matrix was alloyed with Sn-Pb-Cu-Zn-Si-Mo, and selected for alloy fabrication by powder metallurgy method followed by SPS.

Therefore, the purpose of the present work was to investigate the effect of the various constituent’s presence and milling parameters on the microstructure of the Al-based alloys. Besides, as far as we know, attempts to create such microstructures which are liable to rapid friction-induced self-organization, in order to obtain favorable secondary structures that provide the optimal tribological properties, have not been previously investigated in the literature.

2. Materials and Methods

2.1. Materials

For this work, commercial aluminum and zinc (d$_{50}$ = 50 ± 70 µm, 99.9% purity and d$_{50}$ = 40 ± 10 µm, 99.0% purity, respectively,”Plasmotherm” Ltd., Moscow, Russia); tin and lead (d$_{50}$ = 80 ± 10 µm, 99.1% purity and d$_{50}$ = 75 ± 15 µm, 99.95% purity, respectively, JSC “ZMS”, Ryazan, Russia); magnesium and silicon (d$_{50}$ = 80 ± 13 µm, 99.1% purity and d$_{50}$ = 120 ± 30 µm, 99.0% purity, respectively, Yekaterinburg, Russia); copper (d$_{50}$ = 45 ± 13 µm, JSC “Uralelektromed”, Verkhnyaya Pyshma, Russia, 99.5% purity); molybdenum (d$_{50}$ = 5 ± 8 µm, “NPK Metoteknika” Ltd., Moscow, Russia, 99.5% purity) powders were used.

2.2. Compositions Processing

The chemical composition of the experimental aluminum alloy that was investigated in this research is presented in Table 1.

| Chemical Composition, wt. % |  |
|----------------------------|---|
| Sn | Pb | Cu | Zn | Mg | Si | Mo | Al |
| 6  | 2  | 4.3| 1.9| 1.7| 0.45| 1  | Bal. |

Details of powder mixture processing parameters are given in Table 2.

| Simple Mixture 1 (S1) | Simple Mixture 2 (S2) | Simple Mixture 3 (S3) | Compound Mixture | Compound Mixture Number |
|-----------------------|-----------------------|-----------------------|-------------------|-------------------------|
| all elemental powders were milled for 24 h | Sn+Mg+Pb/24 h | Al+Si+Mo/24 h | S1+S2+S3/24 h | M1 |
| Cu+Zn/24 h | Sn+Mg+Pb/24 h | Al+Si+Mo/24 h | S2+S3+Cu+Zn/12 h | M2 |
| --- | Sn+Mg+Pb/48 h | Al+Si+Mo/24 h | S2+S3+Cu+Zn/12 h | M3 |
The paragraph below briefly describes some of the steps in these closely-related processes. During the first step, the M1 mixture was prepared by a conventional processing route, i.e., all raw powders were milled by high-energy milling that was performed using 5 mm diameter hard metal balls and vial for 24 h. A ball-to-powder weight ratio of 10:1 was used. Then the obtained suspension was dried, ground, sieved and sintered.

In further, the milling process was modified. The simple mixtures (S1–S3) were homogenized by milling with WC-Co balls in a tungsten carbide vial during 24 or 48 h, and then dried. The resulting powders were ground in an agate mortar and subsequently passed through a 63 µm sieve. These mixtures of powders were again milled together under the same processing parameters for 24 h. The derived compound mixture was dried and sieved as well, and has been designated as M2. It should be clarified that during the processing of M3 mixture, Cu and Zn powders were added to the compound mixture as separate constituents, and M3 was milled for 12 h. Isopropanol was used in all the studied mixtures as a fluid. Rotational speed was kept constant during all milling operations as 1250 rpm.

2.3. Spark Plasma Sintering (SPS)

Obtained mixtures were placed in a die-punch setup made from isostatic graphite (grade C4, DonCarb Graphite, Rostov, Russia) and consolidated by an H HP D-25 SD spark plasma sintering machine (FCT Systeme GmbH, Rauenstein, Germany) at 500 °C with a heating rate of 100 °C·min⁻¹, under a uniaxial pressure of 50 MPa in a vacuum. The final temperature and pressure were maintained for 3 min. The temperature was controlled during sintering by a pyrometer situated at the top of the machine and focused at the center of the blank (3 mm over the top surface). As-sintered samples were 20 mm in diameter with a thickness of 5 mm.

2.4. Microstructural Characterization

A scanning electron microscope (SEM) VEGA 3 (Tescan, Brno, Czech Republic) was used for the characterization of surfaces polished down to 1 µm. The energy-dispersive X-ray spectroscopy (EDS) spectra were conducted in specific regions at 20 kV voltage and 12 µA beam current for 5 min to obtain well-resolved peaks for the chemical microanalysis of sintered samples.

3. Results and Discussion

Figure 1 shows scanning electron microscope (SEM) micrographs in backscattered electron (BSE) and secondary electron (SE) mode with the energy-dispersive X-ray spectroscopy (EDS) analysis points corresponding to the alloy, which was produced by just milling all powders together for 24 h (M1). In the case of BSE, the lighter elements (low atomic number) appear brighter in the image.

EDS elements concentration data of the M1 alloy are reported in Table 3. Figure 2 shows the SEM-EDS maps of elemental distributions in the M1 alloy.

The structure of this M1 alloy consists of the aluminum particles (Figure 2A) partially jointed together, but in most cases separated by thin walls of various compositions. Several different inclusions of various compositions and shapes between aluminum subgrains were observed. The spectra (3–7, 23, 35 and 36) show that partial saturation of Al-matrix by alloying elements during sintering occurred: Cu–0.21%–1.59%, Zn–0.54%–0.75%, Mg–0.11%–0.36%, Si–0.37%–0.63%.

Two spectra show a small amount of tin, while lead and molybdenum were not detected (spectra 23 and 35). The presence of tin on the aluminum particles is most likely due to the “spreading” effect during grinding and polishing procedures. There was intergranular space, especially over the oxide layer partially filled by tin (Figure 2F) and therefore there are areas with a high content of tin and oxygen (spectra 19, 27, 29, 32, 33 and 41). The results presented show that the tin phase was not only located on the oxide layer but, as the lowest melted element, it reacts with magnesium, lead and zinc. Copper-based solid inclusions have large sizes and are inhomogeneously distributed in the alloy (Figure 2D). The EDS spectra (16, 17, 20, 21 and 40) show these inclusions are based on solid
solutions of \( \theta \)-phase (CuAl\(_2\)) with an excess of aluminum, and additionally doped by Zn, Sn, Mg, and Si. The content of various constituents changes within the following limits: Al: 1.9%–89.13%, Sn up to 80.56%, lead up to 24.87%, magnesium up to 5.08%, silicon up to 5.42%, copper up to 32.64%, zinc, 0.23%–6.05% and Mo up to 94%. Therefore, it may be stated that zinc completely diluted and in various amounts uniformly distributed in all alloy phases (Figure 2E). Meantime the magnesium (Figure 2H) and copper (Figure 2D) are less evenly spaced. For the latter typically, there was the preservation of relatively large inclusions. A large variety of the size of Si (Figure 2B) and Mo (Figure 2C) solid inclusions was observed due to their low chemical reactivity.

The SEM-EDS results indicate that aluminum matrix offers limited mechanical alloying when powders were milled together (spectra 1–5 and 21). The presence of an oxide layer that cannot be removed from the Al particles surface leads to weak adhesion and the mechanical and tribological properties of the alloy will consequently deteriorate. Tin soft phase inclusions do not interact evenly with Pb and Mg to obtain the favorable secondary structures during friction, as it was anticipated (spectra 6, 8, 11, 13, 16 and 17). The solid inclusions with high copper content contain more aluminum than the necessary to achieve the secondary structures and stoichiometric \( \theta \)-phase (CuAl\(_2\)) as well (spectra 14, 15 and 19). Molybdenum particles were inhomogeneously distributed in alloy structure and have a large size. Silicon particles demonstrate a wide variety both in size and uniformity of distribution (spectra 6 and 18).

Therefore, this powder milling method does not allow the homogeneous distribution of alloying elements in the alloy. Consequently, it can be assumed that mechanical properties will be reduced as well. Besides the degradation of mechanical performance, the obtained microstructure does not seem to contribute to the formation of secondary structures on the tribosurfaces and the ability to ensure enhanced tribological properties. Thus, the milling technique must be modified.

Figure 1. Backscattered electron (BSE) mode (A) and SE mode (B) with EDS analysis points SEM images of the M1 alloy.
Table 3. Energy-dispersive X-ray spectroscopy (EDS) chemical identification of elements and their concentration in M1 alloy.

| No. | C    | O    | Mg   | Al   | Si   | Cu   | Zn   | Mo   | Sn   | Pb   |
|-----|------|------|------|------|------|------|------|------|------|------|
| 1   | 3.66 | 5.28 | 0.12 | 89.13| 0.52 | 0.21 | 0.54 | 0    | 0    | 0    |
| 2   | 4.54 | 6.76 | 0.12 | 86.67| 0.55 | 0.26 | 0.75 | 0    | 0    | 0    |
| 3   | 5.39 | 6.98 | 0.21 | 85.87| 0.63 | 0.37 | 0.56 | 0    | 0    | 0    |
| 4   | 4.64 | 5.38 | 0.13 | 88.29| 0.63 | 0.31 | 0.62 | 0    | 0    | 0    |
| 5   | 4.27 | 7.92 | 0.11 | 85.4 | 0.42 | 0.74 | 0.72 | 0    | 0.43 | 0    |
| 6   | 2.25 | 15.36| 0.71 | 37.76| 5.42 | 0.22 | 0.96 | 0    | 29.74| 7.59 |
| 7   | 4.58 | 20.24| 2.79 | 57.91| 0.18 | 0.38 | 0.72 | 0    | 12.54| 0.66 |
| 8   | 2.5  | 25.76| 0.64 | 28.04| 0    | 0.28 | 1.6  | 0    | 26.52| 14.66|
| 9   | 2.69 | 18.77| 0.48 | 58.88| 0    | 0.3  | 0.73 | 0    | 18.15| 0    |
| 10  | 2.93 | 16.62| 0.39 | 64.71| 0.2  | 0.48 | 0.73 | 0    | 13.94| 0    |
| 11  | 6.6  | 39.21| 2.33 | 4.06 | 0.37 | 0.21 | 0.3  | 0.29 | 25.54| 1.1  |
| 12  | 4.85 | 25.85| 0.95 | 48.36| 0    | 1.09 | 0.54 | 0    | 16.99| 1.37 |
| 13  | 6.4  | 18.37| 1.98 | 28.55| 0.15 | 1.32 | 0.75 | 0    | 30.58| 11.89|
| 14  | 8.75 | 5.93 | 1.52 | 42.1 | 0    | 32.64| 6.05 | 0    | 3.01 | 0    |
| 15  | 26.76| 14.56| 0.91 | 39.21| 0.42 | 16.96| 0.33 | 0    | 0.72 | 0    |
| 16  | 5.49 | 21   | 0    | 40.24| 0.18 | 0.45 | 0.59 | 0    | 28.76| 3.3  |
| 17  | 6.5  | 28.83| 5.08 | 1.9  | 0.18 | 0.72 | 3.08 | 0    | 51.61| 2.1  |
| 18  | 6.43 | 3.88 | 0    | 61.91| 0.9  | 25.85| 0.45 | 0    | 0.58 | 0    |
| 19  | 4.2  | 6.21 | 0.3  | 53.04| 0.63 | 21.02| 0.36 | 0    | 14.26| 0    |
| 20  | 16.08| 22.84| 0.8  | 39.31| 0.11 | 1.36 | 0.59 | 0    | 7.35 | 11.42|
| 21  | 8.7  | 6.86 | 0.19 | 82.05| 0.63 | 0.67 | 0.74 | 0    | 0.17 | 0    |
| 22  | 24.54| 28.8 | 3.8  | 4.42 | 0.6  | 0.32 | 0.23 | 0    | 34.38| 0.75 |
| 23  | 11.42| 32.24| 2.31 | 8.76 | 0.64 | 0.4  | 0.36 | 0    | 41.2 | 2.68 |
| 24  | 14.69| 16.84| 0.64 | 61.7 | 0.26 | 2.34 | 0.71 | 0    | 2.39 | 0.43 |
| 25  | 5.06 | 19.79| 3.02 | 2.27 | 0.22 | 0.72 | 0.53 | 0    | 60.63| 7.75 |
| 26  | 2.9  | 28.55| 3.47 | 26.77| 0    | 0    | 0.53 | 0    | 12.91| 24.87|
| 27  | 3.2  | 8.61 | 0.56 | 0.87 | 0    | 0    | 0.58 | 0    | 80.56| 5.61 |
| 28  | 9.4  | 14.98| 0.4  | 69.63| 0.26 | 0.7  | 1.03 | 0    | 3.61 | 0    |
| 29  | 9.14 | 38.6 | 1.11 | 41.48| 0.31 | 0.25 | 0.54 | 0    | 7.33 | 0.66 |
| 30  | 8.92 | 28.02| 4.16 | 1.84 | 0.42 | 3.65 | 2.24 | 0    | 47.14| 3.62 |
| 31  | 7.34 | 15.09| 2.65 | 21.92| 0.2  | 9.8  | 0.38 | 0.33 | 39.99| 2.62 |
| 32  | 4.65 | 21.78| 1.31 | 45.06| 0.09 | 0.3  | 0.55 | 0    | 21.95| 4.32 |
| 33  | 7.54 | 8.38 | 0.36 | 80.15| 0.62 | 1.59 | 0.64 | 0    | 0.42 | 0    |
| 34  | 5.45 | 9.87 | 0.15 | 82.87| 0.37 | 0.59 | 0.71 | 0    | 0    | 0    |
| 35  | 7.62 | 47.03| 5.03 | 12.7 | 0.18 | 0.18 | 0.29 | 0    | 25.7 | 1.12 |
| 36  | 8.37 | 40.69| 2.05 | 35.55| 0.41 | 0.22 | 0.73 | 0    | 10.67| 0.83 |
| 37  | 10.23| 23.48| 0.44 | 58.33| 0.21 | 0.41 | 0.75 | 0    | 5.37 | 0.65 |
| 38  | 7.62 | 4.83 | 0.33 | 60.91| 0.77 | 25.16| 0.38 | 0    | 0    | 0    |
| 39  | 6.76 | 26.99| 3.93 | 13.07| 0.11 | 0.3  | 0.75 | 0    | 43.79| 4.01 |
The microstructure of M1 alloy consists of the aluminum particles (Figure 2A) partially jointed together, but in most cases separated by thin walls of various compositions. Several different inclusions of various compositions and shapes between aluminum subgrains were observed. The spectra (3–7, 23, 35 and 36) show that partial saturation of Al-matrix by alloying elements during sintering occurred: Cu–0.21%–1.59%, Zn–0.54%–0.75%, Mg–0.11%–0.36%, Si–0.37%–0.63%. Two spectra show a small amount of tin, while lead and molybdenum were not detected (spectra 23 and 35). The presence of tin on the aluminum particles is most likely due to the “spreading” effect during grinding and polishing procedures. There was intergranular space, especially over the oxide layer partially filled by tin (Figure 2F) and therefore there are areas with a high content of tin and oxygen (spectra 19, 27, 29, 32, 33 and 41). The results presented show that the tin phase was not only located on the oxide layer but, as the lowest melted element, it reacts with magnesium, lead and zinc.

Copper-based solid inclusions have large sizes and are inhomogeneously distributed in the alloy (Figure 2D). The EDS spectra (16, 17, 20, 21 and 40) show these inclusions are based on solid solutions of \( \theta \)-phase (CuAl\(_2\)) with an excess of aluminum, and additionally doped by Zn, Sn, Mg, and Si. The content of various constituents changes within the following limits: Al: 1.9%–89.13%, Sn up to 80.56%, lead up to 24.87%, magnesium up to 5.08%, silicon up to 5.42%, copper up to 32.64%, zinc, 0.23%–6.05% and Mo up to 94%. Therefore, it may be stated that zinc completely diluted and in various amounts uniformly distributed in all alloy phases (Figure 2E). Meantime the magnesium (Figure 2H) and copper (Figure 2D) are less evenly spaced. For the latter typically, there was the preservation of relatively large inclusions. A large variety of the size of Si (Figure 2B) and Mo (Figure 2C) solid inclusions was observed due to their low chemical reactivity.

The SEM-EDS results indicate that aluminum matrix offers limited mechanical alloying when powders were milled together (spectra 1–5 and 21). The presence of an oxide layer that cannot be removed from the Al particles surface leads to weak adhesion and the mechanical and tribological properties of the alloy will consequently deteriorate. Tin soft phase inclusions do not interact evenly with Pb and Mg to obtain the favorable secondary structures during friction, as it was anticipated (spectra 6, 8, 11, 13, 16 and 17). The solid inclusions with high copper content contain more aluminum than the necessary to achieve the secondary structures and stoichiometric \( \theta \)-phase (CuAl\(_2\)) as well (spectra 14, 15 and 19). Molybdenum particles were inhomogeneously distributed in alloy structure and have a large size. Silicon particles demonstrate a wide variety both in size and uniformity of distribution (spectra 6 and 18).

Therefore, this powder milling method does not allow the homogeneous distribution of alloying elements in the alloy. Consequently, it can be assumed that mechanical properties will be reduced as well. Besides the degradation of mechanical performance, the obtained microstructure does not seem to contribute to the formation of secondary structures on the tribosurfaces and the ability to ensure enhanced tribological properties. Thus, the milling technique must be modified.

The microstructure of M2 alloy produced using the second method of fabrication is presented in Figure 3.

Table 4 below shows the EDS analysis data and the elements concentration of the M2 composition.

**Figure 2.** SEM-EDS elemental distribution maps for aluminum (A), silicon (B), molybdenum (C), copper (D), zinc (E), tin (F), lead (G) and magnesium (H) for M1 alloy.

**Figure 3.** BSE mode (A) and SE mode (B) with EDS analysis points SEM images of the M2 alloy.

Table 4 below shows the EDS analysis data and the elements concentration of the M2 composition.
Table 4. EDS chemical identification of elements and their concentration in M2 alloy.

| No. | Contents of Elements by Weight (%) |
|-----|------------------------------------|
|     | C    | O    | Mg | Al | Si  | Fe | Cu | Zn | Mo | Sn | Pb |
| 1   | 2.62 | 7.83 | 0  | 85.03 | 0.16 | 0 | 0.48 | 0.77 | 0.26 | 1.84 | 1.02 |
| 2   | 3.19 | 7.36 | 0  | 89.06 | 0.17 | 0 | 0.21 | 0   | 0   | 0   | 0   |
| 3   | 3    | 16.16| 0.13| 79.09 | 0   | 0 | 0   | 0.33 | 0.62 | 0.66 |
| 4   | 5.77 | 16.44| 0  | 77.79 | 0   | 0 | 0   | 0   | 0   | 0   |
| 5   | 4.29 | 34.98| 0.77| 20.05 | 0.14 | 0 | 1.63 | 0.84 | 0   | 30.19| 7.11 |
| 6   | 6.62 | 29.6 | 1.27| 6.29  | 0.49 | 0 | 1.44 | 0   | 0   | 51.51| 2.38 |
| 7   | 3.52 | 24.07| 0.64| 15.55 | 0.33 | 0 | 1.17 | 0.6  | 0.31 | 52.02| 1.8  |
| 8   | 3.45 | 5.63 | 2.64| 31.14 | 0.26 | 0 | 49.53| 0.96 | 0   | 6.39 |
| 9   | 3.65 | 1.73 | 0.26| 0.59  | 0   | 0 | 71.91| 21.48| 0   | 0.38 |
| 10  | 7.1  | 4.71 | 0.6 | 1.2   | 0   | 0 | 66.83| 19.56| 0   | 0   |
| 11  | 8.6  | 6.13 | 0.32| 0.94  | 0   | 0 | 50.28| 33.02| 0   | 0.71 |
| 12  | 16.29| 45.16| 2.79| 30.51 | 0.85 | 0 | 0.26 | 0.43 | 1.1  | 1.47 | 1.13 |
| 13  | 2.77 | 11.04| 0  | 86.19 | 0   | 0 | 0   | 0   | 0   | 0   |
| 14  | 4.03 | 28.73| 6.27| 20.62 | 0.38 | 0 | 3.02 | 1.33 | 0   | 29.62| 6   |
| 15  | 4.05 | 3.64 | 0.39| 2.42  | 0   | 0 | 85.32| 3.34 | 0.25 | 0   |
| 16  | 3.7  | 30.29| 3.68| 5.15  | 0.24 | 0 | 0.53 | 0.25 | 0   | 48.44| 7.72 |
| 17  | 3.25 | 20.69| 0.84| 63.02 | 0.15 | 0 | 0.74 | 0.97 | 0.39 | 7.39 | 2.54 |
| 18  | 2.87 | 14.96| 0.29| 75.98 | 0.14 | 0 | 0.74 | 1.06 | 0.35 | 2.18 | 1.44 |
| 19  | 3.65 | 4.71 | 0.08| 90.44 | 0.12 | 0 | 0.3  | 0.7  | 0   | 0   |
| 20  | 5.27 | 24.73| 0.39| 20.18 | 0.14 | 0 | 37.15| 10.53| 0.34 | 0.5  | 0.77 |
| 21  | 3.06 | 17.72| 0.25| 60.18 | 0.15 | 0 | 1.09 | 0.82 | 0.35 | 12.3 | 4.07 |
| 22  | 3.29 | 8.11 | 0.25| 45.95 | 0.19 | 0 | 34.41| 0.96 | 0   | 6.37 | 0.46 |
| 23  | 2.25 | 15.32| 0.72| 12.48 | 0   | 0 | 0.35 | 2.12 | 0   | 57.3 | 9.47 |
| 24  | 9.16 | 28.06| 1.56| 3.77  | 0.16 | 1.24| 0.25 | 0   | 44.54| 7.76 | 3.5  |
| 25  | 3.52 | 1.12 | 0  | 0.58  | 0   | 0 | 77.12| 17.66| 0   | 0   |

The EDS provides the following results:
1. The spectra (1–4, 13, 18 and 19) show that partial saturation of Al-matrix by alloying elements during sintering occurred: Cu: 0.21%–0.74%, Zn: 0.33%–1.06%, Mg: 0.08%–0.29%, Si: 0.12%–0.16%, Mo: 0.26%–0.62%. Clusters of Pb and Sn, elements that in casting do not dissolve in the aluminum matrix, were observed due to mechanical alloying. Thus, this low level of the aluminum matrix alloying is undesirable for the mechanical and tribological properties of journal bearings.

2. Solid inclusions of copper-based exhibit the presence of a broad range of various alloying elements (spectra 8–11, 15, 20 and 25). In most cases, the copper reacted significantly with Zn because these metals were milled together, while in two cases a higher content of aluminum than Zn was observed (spectra 8 and 20). This moves closer to the initial structure of the M2 alloy to being beneficial for the secondary structure of the tribosurface θ-phase that prevails on the friction surface of alloys with a high score and wear resistance. Only in one case reaction with tin or so-called “bronzing” was detected (spectrum 8). Moreover, this analyzed point is aluminum-enriched and low in zinc. The small portion of light alloyed copper, and there are five more metals with a relatively low content of C and O that was found (spectrum 15).

On two separate occasions, the reaction of copper of either high or low concentration with Mo was detected that would suggest randomness of this interaction (spectra 15 and 20).
3. The soft phase is characterized by the higher content of Sn than Pb (spectra 5–7, 14, 16 and 23). It is evident that Sn phases can be easily alloyed, not only Pb but Mg, Zn and Cu as well. In addition, solid inclusions of Si, Mo and alumina can be absorbed by these phases. As a result pseudograins of a babbitt-like structure that should positively affect self-organization during friction were formed.

Figure 4 shows the SEM-EDS maps of elemental distributions in the M2 alloy.

Figure 4. SEM-EDS elemental distribution maps for aluminum (A), silicon (B), molybdenum (C), copper (D), zinc (E), tin (F), lead (G) and magnesium (H) for M2 alloy.

Based on the foregoing and Figure 4, it can be assumed that:

1. Most of the aluminum particles coalescence into larger grains due to losing an oxide layer (Figure 4A), while there are particles covered with this layer.

2. Homogeneous distribution and reducing the size of Si and Mo particles in the alloy matrix confirms the correctness of milling of these powders with aluminum (Figure 4B,C).

3. Cu (Figure 4D) and Zn (Figure 4E) powders formed brass-plated particles through the interaction with each other. On the one hand, this reduced Sn amount of the copper bronzing is to the detriment of the creation of soft phases. On the other side, the interaction of brass-plated particles with aluminum is weak at the second stage of milling. Consequently, an insufficient amount of solid particles θ-phase (CuAl2) required for secondary structures creation, that are responsible for superior tribological properties, were formed. Therefore, the milling time of these powders at the first stage of preparation can be reduced.

4. At the first step of producing Sn (Figure 4F), Pb (Figure 4G) and Mg (Figure 4H) powders are mechanically alloyed each other with the creation of the three-component phases of different compositions. These phases easy absorb solid inclusions of the other phases, primarily Al2O3. However, as would be expected the preparation time of the S2 mixture could be increased.

All of this has a favorable effect on the alloys self-organization during friction and formation of secondary structures on both alloy and steel roller tribosurfaces.

Thus, taking into account the results of M1 and M2 milling techniques on the microstructure of studied alloy, milling parameters were modified. It leads to the next milling procedure. This method is different from the one considered earlier by the refusal to mix Co and Zn powders together (i.e., simple mixture S1). Additionally, the S2 mixture for M3 alloy was milled twice as long as before, because this period provides the mechanical alloying effect of Sn particles with Pb and Mg.

Phases in the Sn-Pb-Mg system were previously found to ensure the maximum level of tribological properties, particularly wear resistance [19,20]. On the one hand, Sn is more doped by Mg and Pb, and the amount of soft lead-based releases was reduced as well. On the other hand, the tin reacts...
less with copper. Meanwhile, magnesium dopes the aluminum matrix less, which prevents excessive increases in alloy hardness.

Also, the milling time of the S3 mixture was not modified. By using such milling parameters, the Al particles were cleaned from the oxide layer as possible. At the same time, these particles were mechanically alloyed by Si, and some particles alloyed by Mo. The size of Si and Mo particles was reduced, and these particles were homogeneously distributed in the alloy matrix. Then, Co and Zn powders, S2 and S3 mixtures, were ball milled for 12 h. As a result, an M3 mixture was obtained.

The microstructure with the EDS spectra of sintered alloy produced by the M3 technique is shown in Figure 5.

Figure 5. SEM BSE mode (A) and SE mode (B) with EDS analysis points images of the M3 alloy.

Studies have shown that aluminum particles united in grains with the optimal size for antifriction material (Figure 6A). Zinc (Figure 6E) and magnesium (Figure 6H) particles, small inclusions of Si (Figure 6B) and Mo (Figure 6C) phases were homogeneously distributed throughout the aluminum matrix. The larger solid inclusions have a copper base (Figure 6D). In the tin-lead phases (Figure 6F,G) the presence of alumina and small inclusions of solid phases have been observed.

Figure 6. SEM-EDS elemental distribution maps for aluminum (A), silicon (B), molybdenum (C), copper (D), zinc (E), tin (F), lead (G) and magnesium (H) for M3 alloy.
Results of EDS analysis with elements concentration of the Al alloy milled by M3 method are shown in Table 5.

| No. | C  | O  | Mg | Al  | Si  | Cu  | Zn  | Mo  | Sn  | Pb  |
|-----|----|----|----|-----|-----|-----|-----|-----|-----|-----|
| 1   | 2.06 | 12.09 | 0.36 | 77.62 | 0.16 | 1.04 | 1.97 | 0.24 | 3.5 | 0.94 |
| 2   | 1.61 | 10.98 | 0.38 | 78.11 | 0.19 | 1.09 | 1.93 | 0   | 4.59 | 1.11 |
| 3   | 1.89 | 8.51  | 0.66 | 81.75 | 0.44 | 0.99 | 2.11 | 0.14 | 2.53 | 1.12 |
| 4   | 1.7 | 5.5 | 0.12 | 88.81 | 0.3 | 1.57 | 2.21 | 0   | 0.29 | 0   |
| 5   | 2.69 | 13.31 | 1.12 | 69.18 | 0.21 | 1.03 | 1.88 | 0 | 4.35 | 6.23 |
| 6   | 1.97 | 13.95 | 0.23 | 78.63 | 0.19 | 1.49 | 1.86 | 0.44 | 0.84 | 0.4 |
| 7   | 2.16 | 5.92 | 0.17 | 85.36 | 0.24 | 0.88 | 2   | 0.15 | 1.45 | 0.47 |
| 8   | 6.88 | 5.02 | 0.09 | 81.64 | 0.2 | 2.43 | 2.16 | 0   | 1.67 | 0   |
| 9   | 2.1 | 6.64 | 0.23 | 80.11 | 0.24 | 5.42 | 1.91 | 0   | 2.48 | 0.87 |
| 10  | 1.84 | 4.81 | 0.28 | 87.57 | 0.26 | 1.06 | 2   | 0   | 1.02 | 1.06 |
| 11  | 1.07 | 1.99 | 0.46 | 42.29 | 0.3 | 43.91 | 0.71 | 0 | 7.45 | 1.82 |
| 12  | 1.91 | 0.95 | 0.4 | 27.71 | 0 | 63.26 | 2.98 | 0 | 2.08 | 0.71 |
| 13  | 1.52 | 2.57 | 1.02 | 42.9 | 0.34 | 46.31 | 0.71 | 0 | 3.45 | 1.17 |
| 14  | 4.11 | 20.02 | 5.24 | 3.14 | 0.3 | 3.58 | 6.66 | 0 | 39.02 | 17.32 |
| 15  | 1.82 | 1.71 | 0.17 | 45.65 | 0.29 | 48.47 | 0.44 | 0 | 1.02 | 0.42 |
| 16  | 1.72 | 6.78 | 0.36 | 32.33 | 2.89 | 38.27 | 1.36 | 0 | 14.05 | 2.24 |
| 17  | 5.08 | 11.73 | 0.8 | 2.01 | 0.28 | 12.07 | 0.84 | 0 | 65.11 | 2.09 |
| 18  | 8.16 | 15.46 | 2.46 | 4.43 | 0.18 | 5.26 | 1.22 | 19.14 | 41.02 | 2.66 |
| 19  | 2.69 | 6.12 | 1.62 | 30.73 | 0.26 | 36.71 | 2.05 | 0 | 11.74 | 8.2 |
| 20  | 2.03 | 2.76 | 0.16 | 44.87 | 0.31 | 46.61 | 0.62 | 0 | 1.78 | 0.87 |
| 21  | 1.88 | 1.56 | 0.31 | 45.08 | 0.33 | 49.66 | 0.66 | 0 | 1.18 | 0 |
| 22  | 2.55 | 3.36 | 0.14 | 35.45 | 0.35 | 39.2 | 0.72 | 0 | 7.41 | 10.82 |
| 23  | 2.05 | 9.31 | 0.16 | 39.4 | 0.17 | 0.98 | 1.28 | 42.86 | 2.08 | 1.7 |
| 24  | 2.44 | 16.23 | 0.5 | 55.04 | 0.18 | 7.92 | 1.51 | 12.11 | 2.63 | 1.46 |
| 25  | 2.58 | 16.66 | 0.21 | 73.41 | 0.16 | 1.63 | 1.96 | 1.56 | 1.32 | 0.49 |
| 26  | 1.86 | 12.44 | 0.17 | 73.35 | 0.19 | 1.49 | 1.71 | 4.77 | 2.04 | 1.97 |

The spectra (1–4 and 6–10) exhibit that partial saturation of Al-matrix by alloying elements during sintering occurred: Cu: 0.99%–5.42%, Zn: 1.86%–2.21%, Mg: 0.09%–0.66%, Si: 0.16%–0.44%, Sn: 0.29%–4.59%, Pb up to 1.12% and Mo up to 0.44%. The aluminum matrix is more homogenous and shows a higher alloying degree compared to other mixtures, and that proved the usefulness of this milling technique. Silicon solid inclusions are so small that determination of their composition, even with the SEM, is not possible. For the large molybdenum-based solid inclusion, which is inside the tin-based phase, 42.86 wt. % of Mo was detected (spectrum 18). Most of the large solid inclusions have a copper base (spectra 11–13, 15, 16, 19, 20 and 22). In seven of the eight spectra the presence of θ-phase (CuAl₂) further alloyed with Zn, Mg, Sn and Pb, was found. Only in one case (spectrum 12), the composition of the inclusion was not close to a stoichiometric due to a lack of aluminum and excess of copper. Perhaps the largest copper particle was not mechanically alloyed with aluminum up to the required level. However, this particle was the only without Si and with the maximum content of oxygen and Zn. In doing so, the inclusions with low copper content (spectra 11, 22, 16 and 19) show higher concentrations of Sn, i.e., a bronzing effect by using tin which is essential for the soft phase in...
Metals 2019, 9, 1164
12 of 14

alloy. It can be assumed that hardness and strength growth may result. Phases with a high content of tin (spectra 14, 17 and 18) are larger than in two previously studied alloys. In addition, some of them were inside of copper inclusions, which is expected to contribute favorably to tribological properties. This effect during the fabrication of other alloys was not observed. The obtained results illustrate the dependence of the inclusions size and tin content in them with the content of most other alloying elements. The larger inclusions with a higher tin content, lower lead, manganese, oxygen and zinc, but more copper. It can be expected that these inclusions that are not available in the alloys produced by other milling techniques are capable of being a trap for the abrasive particles produced during the friction. This is substantiated by the presence of a solid Mo particle (spectrum 18).

4. Conclusions

In this work, the influence of milling parameters on the microstructure of aluminum-based alloys fabricated by SPS was investigated. The following conclusions were obtained from the experimental results:

1. The powder milling method (M1), when all constituents were grouped together in a container and just milled together for 24 h, is not suitable due to the inhomogeneous distribution of alloying elements in the alloy matrix. Additionally, the SEM-EDS study revealed the presence of the oxide layer on the surface of the Al particles, and that cannot be removed during the preparation process. However, if to remove the oxide layer, it should facilitate surface melting on the particle. The solid inclusions with high copper content contain more aluminum than the necessary to achieve the stoichiometric θ-phase (CuAl2). Thus, the milling technique must be modified.

2. The second milling method (M2) proposed was all constituents divided into three groups and each group was milled for 24 h. Then obtained mixtures were grouped together and also milled for 24 h. However, this method has also exposed deficiencies, such as the low level of the aluminum matrix alloying. In addition, an insufficient amount of solid particles θ-phase (CuAl2) that are responsible for superior tribological properties were formed. Therefore, the milling time of Cu and Zn powders at the first stage of preparation should be reduced. On the other hand, the preparation time of the Sn+Mg+Pb mixture could be increased.

3. The last method (M3) implies an increase of milling time for the Sn+Mg+Pb mixture and phase-out of preliminary milling of Cu and Zn powders. According to the study of this method, the alumina matrix is more homogenous and shows a higher alloying degree compared to other milling techniques. In large copper-based solid inclusions the presence of θ-phase (CuAl2) further alloyed with Zn, Mg, Sn and Pb, was found. In addition, some of the phases with a high content of tin were inside of copper inclusions, which is expected to contribute favorably to tribological properties, because these may constitute a trap for the abrasive particles produced during the friction. These microstructure features in alloys obtained by means of M1 and M2 milling methods were not observed. Hence, the microstructure of alloy fabricated by using an M3 milling process should be recognized as the most promising in obtaining the secondary structures on the friction surfaces due to the diversity of the initial structure components. It allows us to suggest that this fabrication method should be used for the milling of powders in these types of alloys.

Further studies are needed for the mechanical and tribological investigation of these alloys. This understanding will enable the processing and designing of the most advantageous properties of these compositions. In addition, the variation in the element’s contents, which affect the microstructural evolution during processing, might also represent the prospective goal for further investigation.

Author Contributions: Conceptualization, A.S. and A.M.; data curation, E.K. and Y.P.; formal analysis, P.P. (Pavel Podrabinnik) and P.P. (Pavel Peretyagin); funding acquisition, P.P. (Pavel Peretyagin); investigation, P.P. (Pavel Podrabinnik) and Y.P.; methodology, A.S. and I.G.; project administration, A.S. and A.M.; resources, I.G. and P.P. (Pavel Peretyagin); software, P.P. (Pavel Podrabinnik) and I.G.; supervision, A.S. and A.M.; validation, Y.P. and E.K.; visualization, P.P. (Pavel Podrabinnik) and E.K.; writing—original draft, A.S. All authors read and approved the final version of the manuscript.
**Funding:** We would like to thank the Ministry of Science and High education of the Russian Federation for supporting this work under the grant number 075-15-2019-1254 with unique identification number RFMEFI57417X0179.

**Conflicts of Interest:** The authors declare no conflict of interest.

**References**

1. Pathak, J.P.; Mohan, S. Tribological behavior of conventional Al–Sn and equivalent Al–Pb alloys under lubrication. *Bull. Mater. Sci.* 2003, 26, 315–320. [CrossRef]
2. Liu, X.; Zeng, M.Q.; Ma, Y.; Zhu, M. Promoting the high load-carrying capability of Al-20 wt.% Sn bearing alloys through creating nanocomposite structure by mechanical alloying. *Wear* 2012, 294–295, 387–394. [CrossRef]
3. Szymczak, T.; Gumienny, G.; Pacyniak, T. E.
4. Kotadia, H.R.; Das, A.; Doernberg, E.; Schmid-Fetzer, R. A comparative study of ternary Al-Sn-Cu alloys for bearing applications. *Appl. Surf. Sci.* 2009, 255, 8202–8206. [CrossRef]
5. Riahi, A.R.; Edrisy, A.; Alpas, A.T. E.
6. Marrocco, T.; Driver, L.C.; Harris, S.J.; McCartney, D.G. Microstructure and properties of thermally sprayed Al-Sn-based alloys for plain bearing applications. *J. Therm. Spray Technol.* 2006, 15, 634–639. [CrossRef]
7. Kotadia, H.R.; Doernberg, E.; Patel, J.B.; Fan, Z.; Schmid-Fetzer, R. Solidification of Al-Sn-Cu immiscible alloys under intense shearing. *Metall. Mater. Trans. A* 2009, 40, 2202–2211. [CrossRef]
8. Kotadia, H.R.; As, A.; Doernberg, E.; Schmid-Fetzer, R. A comparative study of ternary Al-Sn-Cu immiscible alloys prepared by conventional casting and casting under high-intensity ultrasonic irradiation. *Mater. Chem. Phys.* 2011, 131, 241–249. [CrossRef]
9. Pola, A.; Montesano, L.; Gelfi, M.; Roberti, R. Semisolid processing of Al-Sn-Cu alloys for bearing applications. *Sol. St. Phen.* 2013, 192–193, 562–568. [CrossRef]
10. Zhai, W.; Hu, L.; Geng, D.L.; Wei, B. Thermodynamic properties and microstructure evolution of ternary Al–10% Cu–x % Sn immiscible alloys. *J. Alloy. Compd.* 2015, 627, 402–409. [CrossRef]
11. Szymczak, T.; Gumienny, G.; Pacyniak, T. Effect of Vanadium and Molybdenum on the Crystallization, Microstructure and Properties of Hypoeutectic Silumin. *Arch. Foundry Eng.* 2015, 15, 81–86. [CrossRef]
12. Pisarek, B.P. The crystallization of the aluminum bronze with additions of Si, Cr, Mo and/or W. *Arch. Mat. Sci. Eng.* 2007, 28, 461–466.
13. Kuznetsova, E.; Gershman, I.; Mironov, A.; Podrabinnik, P.; Peretyagin, P. The effect of elements of secondary structures on the wear resistance of steel in friction against experimental aluminum alloys for monometallic journal bearings. *Lubricants* 2019, 7, 21. [CrossRef]
14. Smirnov, A.; Peretyagin, P.; Solis, N.; Gershman, I.; Bartolomé, J.F. Wear behavior of graphene-reinforced alumina–silicon carbide whisker nanocomposite. *Nanomaterials* 2019, 9, 151. [CrossRef] [PubMed]
15. Grigoriev, S.; Volosova, M.; Peretyagin, P.; Seleznev, A.; Okunkova, A.; Smirnov, A. The effect of TiC additive on mechanical and electrical properties of Al2O3 ceramic. *Appl. Sci.* 2018, 8, 2385. [CrossRef]
16. Smirnov, A.; Seleznev, A.; Solis, N.; Pritinskii, Y.; Peretyagin, P.; Bartolomé, J.F. The influence of wire electrical discharge machining cutting parameters on the surface roughness and flexural strength of ZrO2/TiN ceramic nanocomposites obtained by spark plasma sintering. *Nanomaterials* 2019, 9, 1391. [CrossRef]
17. Smirnov, A.; Beltrán, J.I.; Rodriguez-Suarez, T.; Pecharromán, C.; Muñoz, M.C.; Moya, J.S.; Bartolomé, J.F. Unprecedented simultaneous enhancement in flaw tolerance and fatigue resistance of zirconia–Ta composites. *Sci. Rep.* 2017, 7, 44922. [CrossRef]
18. Smirnov, A.; Peretyagin, P.; Bartolomé, J.F. Processing and mechanical properties of new hierarchical metal-graphene flakes reinforced ceramic matrix composites. *J. Eur. Ceram. Soc.* 2019, 39, 3491–3497. [CrossRef]
19. Podrabinnik, P.A.; Gershman, I.S.; Mironov, A.E.; Kuznetsova, E.V.; Peretyagin, P.Y. Mechanisms involved in the formation of secondary structures on the friction surface of experimental aluminum alloys for monometallic journal bearings. *Lubricants* 2018, 6, 104. [CrossRef]
20. Mironov, A.E.; Gershman, I.S.; Gershman, E.I.; Zheleznov, M.M. Relationship between the tribological properties of experimental aluminum alloys and their chemical composition. *J. FRICT. WEAR* **2017**, *38*, 87–91. [CrossRef]

21. Mironov, A.E.; Gershman, I.S.; Gershman, E.I. Influence of tin on the tribotechnical properties of complex antifriction aluminum alloys. *J. FRICT. WEAR* **2018**, *39*, 394–399. [CrossRef]

22. Podrabinnik, P.A.; Mironov, A.E.; Gershman, I.S. The influence of secondary structures on wear resistance of experimental aluminum alloys for monometallic slide bearings. *Mat. Today Proceed.* **2019**, *11*, 175–180. [CrossRef]

23. Mironov, A.E.; Gershman, I.S.; Ovechkin, A.V.; Gershman, E.I. Comparison of scoring resistance of new antifriction aluminum alloys and tradition al antifriction bronze. *J. FRICT. WEAR* **2015**, *36*, 257–261. [CrossRef]