Abnormal texture evolution of accumulative roll bonded Al–Cu by adding alumina particles

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ABSTRACT

Evolution of the texture of an Al–Cu composite after addition of alumina particles was investigated in details using X-Ray Diffraction (XRD) method. Composite samples were produced by the Accumulative Roll Bonding (ARB) and alumina particles were uniformly added to the composite using anodizing technique. The process was accompanied by cutting, stacking and rolling the lamellar materials up to seven times. In early stage of ARB, a shear texture close to the Rotated Cube formed for the Al side. This texture development was also confirmed by microstructural investigation on shear bands during deformation employing optical and scanning electron microscopy. In the last two cycles, a nearly random orientation of grains was developed. For the Cu side, however, the recrystallized Cube texture component was prominent for early ARB cycles due to the recrystallization phenomenon. Moreover, a mixture of rolling and recrystallized texture was detected for the Cu side at higher ARB cycles. Furthermore, evaluation of the grain boundaries showed an increase in the number of high angle grain boundaries for the final ARB cycles in both Al and Cu sides, confirming the occurrence of grain refinement during ARB. Localized particle deformation zone (PDZ) around alumina particles was also proposed to have a contribution to textural evolution of the composite.

1. Introduction

Severe Plastic Deformation (SPD) processes are promising through making fine-grained bulk metallic materials [1, 2]. Among many SPD grain refinement processes, Accumulative Roll Bonding (ARB) is one of the most prominent procedures with a wide variety of applications, gaining a great deal of attraction in recent decades [3, 4, 5, 6, 7, 8, 9, 10, 11, 12]. The process shows an increasing attention owing to its feasibility of producing one main metallic matrix [6, 13] and two or more different metallic matrix [14, 15, 16, 17, 18, 19, 20] composites. Moreover, addition of nonmetallic particles (specifically ceramics) has been reported by several researcher groups [4, 10, 21]. For instance, some researchers [22] studied the influence of adding SiC particles on the microstructural and mechanical properties of aluminum matrix using ARB process. The ceramic particles could induce changes in properties due to their stiff nature [22]. Authors of the present paper have also shown that addition of alumina particles can change the composite properties of Al–Cu composites effectively [23]. Unfortunately, in the open literature, a limited number of studies are available about the texture variations by adding particles to the ARBed metalmatrix composites. In one study [24], researchers showed that adding alumina can have an impact on the formation of texture components in the Al-based composites. Another study on steel-based nanocomposite showed that addition of SiC nanoparticles resulted in the variation of texture components compared to that of pure steel. This change in texture development was ascribed to the presence of nanoparticles in the matrix of the composite [21]. To date, the effect of adding ceramic particles on the texture development in two or more different metal matrix composites has not yet been addressed adequately. Therefore, in the present paper, we determine the effect of addition of alumina particles on the texture evolution in Al–Cu dissimilar metal matrix composite using XRD methods, which is aligned with our other research on producing Al–Cu composites (e.g. [23] and [25]). The main idea of this paper is to obtain a better understanding of texture formation in ARBed and FCC metal matrix composites.

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2. Experimental procedure

In this study, as the first step, Cu and Al sheets (as specified in Table 1) were annealed for 1 h at 480 °C and 370 °C, respectively, accompanied by air-cooling to ambient temperature. Secondly, to obtain alumina particles, we used anodizing process: Al sheets were polished by NaOH and then anodized by means of a 15 wt.% Sulfuric acid solution at a voltage of about 15 V and for a soaking time of 30 min. In order to maintain the temperature of the solution to about 10 °C the solution was stirred constantly. The thickness of the oxid film layer was evaluated to be 20 ± 0.2 μm by metallographic procedure with at least 20 measurements for each specimen. The alumina covered the Al sheets in the form of a continuous film and showed no considerable morphology. Finally, the ARB process was conducted in two steps: Al sheet (20 cm (L) × 5 cm (W) × 0.3 mm (T)) was put among two Cu sheets (20 cm (L) × 5 cm (W) × 0.5 mm (T)) to produce an Al-75.5% Cu-3% alumina composite. Afterwards, the intimate strips were joined by rolling to achieve 50% reduction in thickness. The sheets were then cut in to the same size, and the two cut strips were undergone similar surface preparation (stacking and brushing) to be roll bonded for an extra thickness reduction (50%).

The final step was conducted at ambient temperature up to 7 times to gain a uniform composite. A schematic illustration of the procedure is observed in Figure 1. The selected Al-75.5% Cu-3% alumina specimens were wire cut at variety (1st, 3rd, 5th and 7th) ARB cycles for textural evaluations. For detailed microstructural investigation, another composite (Al-21.5% Cu-3% alumina) using different metal thicknesses (1 mm for Al and 0.15 mm for Cu) but the same width and length were produced by the abovementioned procedure.

Texture investigations was applied along the quarter (1/4) thickness on the RD (rolling direction)-TD (transverse direction) plane (an area of 25 × 15 mm²), as shown schematically in Figure 2. Then, different incomplete Pole Figures (PFs), (111), (311), and (220), were obtained assisting X-ray Diffraction (XRD) method. The TexTools software was used to measure the Orientation Distribution Function (ODF) which is located in the Euler space (0° ≤ ϕ₁, Φ, ϕ₂ ≤ 90°). The result of texture measurements was analyzed after different processing cycles for both layers (Al and Cu) and misorientation angle distribution was measured by the same software.

Moreover, shear punch test was applied on the composite by a Houndsfield H50KS testing machine on different ARB cycles and average of three shear forces (N) were collected and reported. The punch and the die used in the present study had a diameter of 3 mm and 4 mm, respectively. The test was conducted on three random places perpendicular to the RD-TD plane of the ARBed composites at a testing speed of 0.5 mm/min. The fractured surfaces (after the tensile test) were evaluated using Scanning Electron Microscopy (SEM) (Philips, the Netherlands) and the microstructural features were studied by optical microscopy (OM) on different rolling directions.

Table 1. Specification of initial Al and Cu.

|   | Al (%) | Fe (%) | Si (%) | Ti (%) | Mn (%) | Sb (%) | V (ppm) | Ni (ppm) | Sn (ppm) | Zn (ppm) | Cu (ppm) |
|---|--------|--------|--------|--------|--------|--------|---------|----------|----------|----------|----------|
| Al | 99.64  | 0.245  | 0.043  | 0.019  | 0.015  | 0.012  | 100     | 63       | 4        | 0        | 0        |
| Cu | 99.86  | 0.060  | 0.0206 | 0.017  | 84     | 73     | 34      | 34       | 30       | 20       | 20       |

Figure 1. Schematic of the ARB process used in the present study.
3. Results and discussion

The ODFs of the Al and Cu sides of the produced composite are presented in Figures 3 and 4, respectively. As-received Al shows a conventional FCC type texture evolution around β fiber, and Rotated Cube is the major component. In the 1st ARB cycle, for both Al and Cu, a nearly random distribution of grain orientations is observed. This is confirmed by comparing the textural components in this cycle with the next ARB cycles, (a nearly random distribution of the components is obvious). This is related to the fact that in this cycle, annealed grains with different orientations start to respond to the rolling force and there is no sufficient time and deformation to form strong and specific texture components.

Figure 3. ODF plots of Al side for a) as received Al, b) 1st, c) 3rd, d) 5th, e) 7th ARB cycles and f) schematic figure of the major textural components in metals with FCC structure.
Further ARB process, however, leads to the development of textural components through rotation of specific plane-direction sets as a response to the rolling force. For instance, for Al side, the major texture components after the 3rd cycle are found to be Rotated Cube, Dillamore, Copper, Brass, and Goss (tabulated in Table 2). In addition, by further ARB of the specimens, it is seen that identical but weaker texture components form in the 7th cycle in comparison to that of the 3rd cycle (Figure 3).

Figure 4. ODF plots of Cu side for a) as received, b) 1st, c) 3rd, d) 5th, e) 7th ARB cycles and f) schematic figure of the major textural components in metals with FCC structure.

ODF maps in Figure 4 for the Cu side after a variety of ARB cycles show that the dominant texture components are Cube (9 × R in the 3rd cycle), Brass and Copper. In addition, as can be observed in Figure 4, the maximum intensity of texture components occurs for the 3rd cycle. For the 5th cycle, intensity of Cube, Copper, Brass, Dillamore and S decrease noticeably. In the final cycle, intensity of Cube decreases (similar to previous cycles), however, intensity of Brass and Copper components increase slightly and other components remained identical (Figure 4).
Based on the literature, we may expect a notable Rotated Cube for both Al and Cu sides along with other components. According to the literature, the evolution of texture is different in Al and Cu sides. In Al side, firstly a significant Rotated Cube develops. By increasing the ARB cycles, this component is weakened dramatically. For the Cu side, a strong Cube component is found at the 3rd ARB cycle. This component remains one of the major ones for the last ARB cycle together with Copper and Brass components.

Textural variations are presented in Figure 6 for each component in both Al and Cu sides. Major components are distinguishable by comparing different components. For Al side, after 7th cycle, a random-like orientation is observed (all component with similar intensity of about 1xR). For Cu side, however, Cube and β-fiber are major components.

For metals with FCC structure such as Al and Cu, a well-developed β-fiber texture has been observed around Brass and Copper [26]. In some studies, researchers argued that as a component was weakened, other components were strengthened [1, 27]. A group of scholars announced that a strong Rotated Cube remained as the major component for Al-alumina composite with various amounts of alumina during ARB process [24]. In another research on ARBed commercial Cu, a shift from the Cube to Rotated Cube was observed after the second ARB cycles [28]. Based on the literature, we may expect a notable Rotated Cube for both Al and Cu sides along with other β-fiber components such as Brass and Copper, and a negligible Cube texture in Cu side at higher ARB cycles. In the present study, in contrast to the above-mentioned studies, we observed an abnormal texture behavior: in some cases, constant fraction of textural intensities was gained by preceding ARB cycles (for instance, Rotated Cube, Copper and Brass for Al side after 5th and 7th cycles); in some other cases, decrease in all components (such as Cube, Brass, Copper, Dillamore and S for Cu side for 5th ARB cycle) was observed. Moreover, a component that was believed to be the major one, i.e. Rotated Cube in Cu, was absent. On top of that, a noticeable recrystallized Cube intensity after the 3rd cycle was found in the present process (conducted at ambient temperature). These are unexpected and interesting findings, which are the main subject of the present study.

The results obtained in the present study will be justified and discussed in the following section. The first point to mention is that resistance between roller and the metal surface may assist transforming of the annealing texture to Rotated Cube [26, 27]. In the present study, Cu sheets were in direct touch with the surface of the rollers. Consequently, Rotated Cube component was presumed to be observed. However, no Rotated Cube was seen in ODF plots (Figure 4) and pole figures (Figure 5). The absence of this component in Cu side suggests that shear texture formed in the surface could have been rotated towards rolling texture in the middle of the strips during ARB process. Consequently, no sign of shear texture was detected in the area where XRD was conducted (quarter thickness). Nevertheless, formation of Rotated Cube was observed in as-received Al specimen and ARBed Al after 3rd cycle (Figure 6). Presence of this texture was ascribed to the evolution of shear bands in the ARBed composites [19]. At the 3rd cycle, owing to the shear bands, a texture (mainly) around Rotated Cube was found (Figures 3 and 6). In order to assert the evolution of the shear bands, OM was used and the fracture surface of the composite material was examined by SEM after tensile test. As it is seen in Figure 7a,b, the Cu layers are fractured due to the formation of shear bands. Effect of shear bands on the re-distribution of Cu layers is shown by dashed lines in Figure 7a,b. In addition, shear dimples that are rotated in a specific direction can be observed after tensile test (Figure 7c). This rotation in dimples is mainly related to the existence of shear bands in the material, turning the dimples in the direction of shear. This behavior in dimples has been recorded in similar materials subjected to the ARB process [10, 29]. Shear component, however, was weakened from 5th cycle onward for Al side. It is believed that this shear texture may transform to a rolling texture (i.e., Copper and Brass) by crystal rotation at the middle of the specimen [30]. Even these two components (Copper and Brass) slightly changed for elevated cycles at Al side (Figure 6).

Moreover, twinning by deformation is reported to be another source of textural evolution at Cu side [27]. It is believed that deformation may induce twinning, which is considered as an important factor in transforming the Copper-type to Brass-type component. However, as indicated in Figure 6, no transition from Copper to Brass was observed. As it was mentioned previously, texture components obtained in the present study are not in good agreement with previous results reported by other research groups (e.g., [19, 24, 26, 27]).

As stated above, one of the most important observations in the study was the intensity of the texture components, which was not constant and in most cases, was decreased by preceding the plastic deformation process.

**Table 2. Dominant texture components and corresponding orientations observed in the present study.**

| Component | Orientation |
|-----------|-------------|
| Copper    | (112) <111> |
| Brass     | (011) <211> |
| Dillamore | (4 4 1 1) <11 1 1 8> |
| Rotated Cube | (001) <110> |
| Goss      | (011) <100> |
| Cube      | (001) <100> |
| S         | (123) <634> |

Pole Figures (PFs) of Al and Cu sides are presented in Figure 5. Al and Cu are both classified as face-centered cubic (FCC) metals and undergo a similar process simultaneously (i.e., ARB). However, as can be observed in PFs (Figure 5), the evolution of texture is different in Al and Cu sides. In Al side, firstly a significant Rotated Cube develops. By increasing the ARB cycles, this component is weakened dramatically. For the Cu side, a strong Cube component is found at the 3rd ARB cycle. This component remains one of the major ones for the last ARB cycle together with Copper and Brass components.

![Figure 5. Pole figures of Al and Cu side at different ARB cycles.](image-url)
Figure 6. Maximum intensity distribution of different components during ARB process for both the Al and Cu.
During ARB, density of dislocations is significantly increased within individual grains, resulting in work hardening of the material [1]. This work hardening, in turn, may affect the mechanical properties of the material. As it is seen in Figure 8, by increasing ARB cycle to 7 (except for some cycles), the maximum shear force (N) of the material shows an increasing trend. For the 4th and 5th ARB cycles, however, a decrease in shear force is seen, which is correlated with the domination of the Al as the softer metal in the matrix (this phenomenon is discussed in detail in our previous study [23]). Thereafter, the higher ARB cycles led to more dislocation formation and work hardening. As a result, shear force increased again for the two last ARB cycles (Figure 8).

During ARB and under special circumstances (such as movement of the dislocations in specific crystallographic planes as a result of the applied force), dislocations may arrange in a special configuration to form Dense Dislocation Walls (DDWs). Increasing the level of deformation may lead to a significant increase of DDWs formation with misorientations more than 15°, which are classified as High Angle Grain Boundaries (HAGBs) [1]. This process can be considered as grain fragmentation, in which a large population of lamellar-shape sub-grains (LAGBs and HAGBs) forms in a single primary grain. In order to verify formation of such HAGBs in the present study, diagrams consist of misorientation and their corresponding fraction are plotted in Figure 9 for Al and Cu sides. As can be seen in Figure 9, by further ARB cycles, volume fraction of HAGBs is increased (peak shifting towards right) through grain fragmentation. These observations could confirm the above process that we discussed. A schematic of the proposed grain refinement process is presented in Figure 10.

To discuss the abnormal observations in the present study (i.e., reducing the intensity of some textural components such as Copper, Brass and Cube for Cu side after 5th cycle, and decrease in the intensity of almost all components for Al side after last ARB cycle), we need to take other factors into consideration.

In the metals subjected to SPD, formation of dislocations results in the instability of the material. This is suggested that recovery and recrystallization processes are required to decrease the internal energy of the system and return it to a lower and stable level. The determining factor in the recovery process is the Stacking Fault Energy (SFE), which in the case of our composite material, is 78 mJm⁻² for Cu and 166 mJm⁻² for Al. In general, in high SFE materials such as Al, recovery may facilitate dislocation climb and cross slip. However, in low SFE materials (such as Copper) recovery process is difficult, giving rise to the recrystallization and formation of dislocation-free recrystallized individual grain due to the accumulation of stored energy during ARB process [27]. Moreover, the ARB process was done at room temperature, meaning that dislocations can respond only to the applied force during the rolling process. Therefore, the reduction in the texture intensity suggests that recrystallization process might be considered as an important process in lowering the intensity of textural components during cold rolling. However, conventional recrystallization consists of nucleation and growth steps that needed driving force (i.e., heat), while the Al-Cu-alumina composite was produced at ambient temperature. It is important to note that in the ARB process, dislocations are able to move easily during the rolling (due to high SFE) and rearrange themselves to form DDWs with misorientations above 15 degrees at Al side (HAGB). This process is accompanied by movement and domination of newly formed high angle grain boundaries, which is also known as recrystallization (with no nucleation of dislocation-free grains). Consequently, the Al may experience grain fragmentation as a result of this process, owing to great population of the dislocations presented in the room temperature processed ARB (These dislocations are mainly populated in the grain boundaries). It should be added that recovery, anneihilation of dislocations and recrystallization occur at the same time, making it hard to separate them in deformed materials. In the present study, however, the recrystallization, due to high stored energy, may compete with recovery process and give rise to formation of HAGB (as shown in Figure 9).

In this regard, two types of recrystallizations are suggested, namely Continuous Recrystallization (CRX) and Discontinuous Recrystallization (DRX). CRX process is normally observed in materials with higher SFE, e.g. Al in the present study. This CRX may take place in the whole material without a classical nucleation step (only movement of grain boundaries are involved), while DRX comprises of nucleation and growth steps in medium and lower SFE materials (such as Cu in the this study). Observation of a strong Cube component in Figures 5 and 6 could be a result of new fine grains formation with recrystallized Cube texture in Cu side through DRX. This is because medium- SFE materials (Cu) may inhibit recovery and promote recrystallization. Moreover, the ambient temperature process (ARB) may have resulted in the storage of a high level of energy in the material. This stored energy then may be considered as the driving force for the discontinues recrystallization process. In addition, due to the fact that these processes were occurred during the rolling, they were considered as “dynamic recrystallization” in some studies [2, 26]. Therefore, the grain refinement process in materials that
undergone SPD may be the key reason to explain the decrease in the texture components’ intensity (Figure 6) at higher ARB cycles, as compared to those of primary cycles.

To evaluate the influence of alumina particles in the texture evolution, we can also propose Particle Deformation Zone (PDZ) phenomenon [1]. This explains that in the presence of stiff particles such as alumina, the regions around the particles may experience a rotational motion, as it is shown schematically in Figure 11. This may encourage the abnormal texture formation [1]. A weak texture formation at high cycles in Al matrix can also be attributed to the rotation of the lattice around alumina particles. Presence of higher volume fraction of alumina particles in the metal matrix may give rise to the rotation of the lattice. Consequently, a complex texture forms, as can be seen in Figures 5 and 6. This process is accelerated when there is sufficient volume fraction of stiff particles in the metal matrix. As a matter of fact, the effect of PDZ can be neglected if there is not considerable number of particles in the matrix. Closely spaced particles in the matrix may lead to more rotation (as illustrated in Figure 11 b,c), giving rise to lower level of texture component intensities. In the present study, the distance between the particles is close enough to facilitate PDZs (Figure 12), which is in good agreement with the related Al side texture evolution (Figure 6). It is noteworthy to mention that by noticing the size of alumina particles in Figure 12, it is evident that higher ARB cycles resulted in more fracturing of alumina particles and distribution of particles took place more uniformly throughout the specimen. Therefore, in higher ARB cycles, it is expected that PDZ should be more effective as it is concluded from Pole Figures and ODF plots of Al side at 5th and 7th cycles (Figures 3 and 5).

Figure 9. Misorientation angle distribution versus number of cycles and their intensities for ARBed Al–Cu composites.
Another point that can be considered is the effect of particles on the Cu texture evolution. In a similar work conducted on the ARB and folding of Al–Cu composite (with no addition of any particles) [26], formation of the Cube component was observed at higher ARBF cycles. Moreover, in our previous study on the ARBed Al–Cu without addition of particles, the intensity of the Cube was not as strong as what was seen in the present study [2]. Considering Al–Cu-alumina texture (Figure 6), however, the Cube component was found as the strongest texture component at the early cycles of ARB for the Cu side. This explains that the particles may act as preferred sites for nucleation of the grains with Cube texture, and the energy stored to initiate the recrystallization at Cu side may increase due to the presence of alumina fragments.

Considering the results of the present study and related discussion, it can be proposed that a combination of material's characteristics (i.e., different SFE), rolling texture, shear texture, recrystallization (DRX and CRX) and PDZ has contributed to the abnormal textural evolution. In the primary cycle of the process, such as 3rd cycle for Al, turning of shear texture to rolling texture was prominent because as the Rotated Cube were decreased (as compared to as-received Al), the Dillamore, Brass and Copper components were increased (Figure 6). This means that Rotated Cube component has rotated towards rolling texture. Nevertheless, for the higher cycles (i.e., 5th and 7th), the effect of particles (PDZ) as well as fragmentation of grains governed the texture evolution because the intensity of the most components remained around 1.1 \times R; as can be seen in Figure 6, there is no significant change in the components' intensity for the two last ARB cycles in Al side is observed. However, for Cu side (Figures 4 and 6), a noticeable decrease in the intensity of all components is observed for 5th cycle. This may be related to the discontinues recrystallization process by which new grains form (increased misorientation shown in Figure 9). In this cycle,
formation of small individual recrystallized grains (due to the stored energy and the effect of alumina particles) might result in the lowering the intensity of the components (note that growth of Cube grains may be ceased since there is not sufficient driving force at room temperature). Nonetheless, a slight increase in the intensity of different components for Cu side in the last cycle showed that the Cube texture, formed at primary cycles, rotated towards rolling direction with Copper and Brass as other dominant components.

Figure 12. Microstructure of the Al-75.5% Cu- 3% alumina composite with alumina particles a) after 5th, and b) 7th ARB cycles.

Figure 13. A comparison of the different components/intensity involve in the present study to that of without any particles [2].
It is worthy to compare the texture results obtained from Al-Cu-alumina, with the similar composite with no addition of alumina particles from our previous study [2]. The results from both studies are summarized in Figure 13. As it is seen, for the Al side, Cube and Rotated Cube components are almost similar to those of our previous study. However, Dillamore, Copper, Brass, S and Goss components show a remarkable difference. These components (except for the Goss) are clearly stronger as compared to those of Al-Cu-alumina composite. For the Cu side, Dillamore, Copper, Brass, S and Goss possess a nearly similar trend as compared to those of Al-Cu-alumina. For the Cube and Rotated Cube components, changes are considerable. For Cube component, a sudden increase is observed as compared to similar component for Al-Cu composite. For the Rotated Cube, the overall intensity of the component is higher for alumina-containing composite.

For the Goss component, it is seen that presence of alumina particles at Al side results in an increase in its intensity after a gradual decrease. It is claimed that this component is a result of recrystallization at primary annealing stage [30]. In contrast to previous findings, it is observed that existence of alumina particles cannot be neglected specifically at the third cycle for Al. These findings may cast doubt on previous claims about the Goss component formation. It seems that during continues recrystallization (CRX), the Goss component development is preferred when alumina particles are present in the matrix. However, formation mechanism of the Goss component needs further investigation.

The effect of the particles on texture evolution is noticeable as we compare both composites (Figure 13). It seems that addition of extra particles to the matrix of the composite attenuates the rolling texture in the matrix (Al side). For the Cu side, however, particles are considered to be effective on the Cube component. It is suggested that the particles may facilitate the domination of the Cube component and may stabilize this component up to the last ARB cycles, as compared with alumina-free Al-Cu composite. For the Rotated Cube, shear deformation may be strengthened when alumina particles exist. As a result, shear texture at Cu side for alumina-containing composite shows a slight increase at the early ARB cycles.

4. Conclusions

Texture evolution in Al-Cu-alumina is investigated using XRD method. An abnormal texture behaviour is observed for Al and Cu sides at different ARB cycles. The main findings of the present study can be outlined as follows:

- For Al side, a shear texture (i.e., Rotated Cube) forms as the dominant component at early cycles of ARB. However, this Rotated Cube shows a very low intensity (nearly random texture) at elevated cycles.
- For the Cu side, a strong Cube texture is firstly observed maybe as a result of recrystallization process induced by high stored energy and presence of alumina particles. However, the intensity of this Cube component decreases for the higher cycles.
- Abnormal texture evolution is mainly related to a combination of several phenomena: shear texture formation (shear bands induced by rollers), rolling texture (through rotation of components during rolling process), formation of fine grains (due to grain fragmentation and recrystallization) and existence of alumina particles (PDZ).
- In Al side, at higher ARB cycles, formation of refined grains (CRX) could be the main factor that governs texture evolution.
- In Cu side, the recrystallized Cube grains during DRX which is dominant at earlier cycles of the ARB, rotated toward rolling texture at higher cycles, and Copper and Brass components become other major components.
- Localized PDZs can be considered as an additional factor in texture evolution when a significant fraction of the particles is distributed in the matrix.

Declarations

Author contribution statement

Vahid Yousefi Mehr: performed the experiments; analyzed and interpreted the data; wrote the paper.
Mohammad Reza Toroghinejad, Jerzy A. Szpunar: conceived and designed the experiments; contributed reagents, materials, analysis tools or data.
Ahmad Rezaeian: conceived and designed the experiments; wrote the paper.
Hamed Asgari: performed the experiments; contributed reagents, materials, analysis tools or data.

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Data availability statement

Data included in article/supplementary material/referenced in article.

Declaration of interests statement

The authors declare no conflict of interest.

Additional information

No additional information is available for this paper.

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