Experimental and numerical investigation of severe plastic deformation of copper sheets processed by modified-constrained studded pressing

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Keywords: modified-constrained studded pressing, finite element simulation, grain size, penetration test, residual stress, crack

Abstract
In this study, mechanical behavior and microstructural evolution of copper sheets, produced by modified-constrained studded pressing (modified-CSP) as a severe plastic deformation (SPD) method, were investigated by experimental and numerical methods. A finite element model (FEM) was established to analyze distribution of equivalent total strain, distribution of equivalent plastic strain and residual stress for the complete first pass of the process. Local stress concentration was predicted by FEM. Also, copper sheets were deformed by the modified-CSP from the first to the tenth passes. The distribution of residual stress on the surface of the copper sheets was measured at each step of the process. Crack initiation and propagation were investigated by the non-destructive penetration test (PT). The residual stresses were predicted by X-ray diffraction (XRD) via sin²Ψ method. The residual stresses were determined for the annealed and the first pass samples, +128.2 and +80.4 MPa, respectively. The maximum compression residual stress was −62.5 MPa for the tenth pass sample. The microstructural evolution including grain size and dislocation density of samples during the process were investigated by XRD pattern analysis. The average grain size had a significant decrease from ∼35 μm to ∼76 nm for annealed sample and tenth pass of modified-CSP, respectively.

1. Introduction

In the severe plastic deformation (SPD) methods, ultra-fine grained (UFG) substructures are obtained by imposing a large amount of strain on coarse-grained metallic materials [1]. A key feature of SPD processes is the nearly constant dimensions of samples. Thus, SPD processes can be carried out repeatedly till crack initiation. Sheets are one of the most important industrial products made by SPD method. The constrained groove pressing (CGP) method was introduced to produce UFG or nanostructure metallic sheets. In the CGP, there are numerous parameters such as sheet thickness, friction coefficient, and die profile (groove width, groove angle). The die profile, among all, is an important parameter in the CGP. So far, the CGP process has been presented in different die profiles. Thangapandian et al [2] presented three different die profiles namely V-shaped, Flat- and semi-circular grooves. Moreover, Torkestani and Dashtbayazi [3] suggested a new method for dies with two orthogonal grooves called as constrained studded pressing (CSP). Elizalde et al [4] proposed a new die profile which was based on two intercepting sinusoidal profiles.

The die profile is an important parameter which causes surface cracks in the sheets. A great deal of research has been carried out on the relation of die profile and surface crack initiation. The die profile in CGP process was divided by Yoon et al [5] into three regions: undeformed area, shear area, and interface between the undeformed and shear region. Plastic deformation occurs mostly in the shear region, while the undeformed area usually accumulates a small strain during deformation. Simultaneously, the interface region makes up stretching and bending deformation. In this way, the non-uniform distribution of plastic strain at the interface region experiences different amounts of strain during the CGP process. Therefore, strain and stress concentration can cause crack initiation in the interface region. Wang et al [6] investigated the effect of die design on surface cracks.
of commercially pure nickel sheets in the CGP method assuming stress concentration and work hardening make micro-cracks at sheet surface. Besides, it was observed that in comparison with the pure shear mode, the bending deformation mode has a significant role in creating micro-cracks at sheet surface. Similarly, according to Wang et al [7], the interface region in CGP method builds a higher cracking tendency in Cu–Zn alloy sheets. Therefore, considering the die design, one of the most important points in the CGP and the CSP processes is locating the cracks in the sheet through experimental methods.

Moreover, during sheet metal forming, residual stress is considered as a key parameter affecting fracture strength. Die profile which is also another important parameter has a significant effect on stress distribution that can be measured. Since the sheet surface experiences more intense plastic deformation than interior, the residual stress values vary greatly in surface and interior of the sheet [8]. Gu et al [9] proposed a new severe plastic deformation process, called ring wave repeated groove pressing, and concluded that the created residual stresses could significantly improve fracture strength. Nazari et al [10] investigated the effect of microstructure on residual stresses of ultrafine-grained copper sheets which were manufactured by CGP process. According to their report, microstructure parameters have a significant effect on residual stress. Despite simultaneous increase in the strain and the number of passes, a considerable decrease is reported in residual stress and grains’ size representing a direct relationship between equivalent strain and residual stress. An important gap in previous research is lack of studies on relations between residual stress and its effect on crack initiation.

Noteworthy, a number of numerical investigations were made on die profile in the CGP method. Evaluating the effect of die profile (width and angle of the die groove) on strain distribution for the CGP process through the commercial software DEFORM-3D, Wang et al [7] reported that due to presence of the interface region, there is a periodic fluctuation in strain distribution. Thus, the interface region has an important role on strain distribution and mechanical properties. Wu et al [11] evaluated the effect of CGP process on distribution of Vickers microhardness, and used inhomogeneity factor (IF) as a measure to characterize the hardness uniformity of the sheet quantitatively. Further, they presented that in the first pass, compared with the annealed sample, there is a greater IF value for Cu–Zr alloy sheet which stands for reduced uniformity of hardness. This phenomenon is because of the interface region of the CGP process that is less influenced by shear deformation with a lower hardness than the rest of the area. Borhani and Javanroodi [12] experimentally investigated the effect of die geometry with a rubber pad using the finite element method (FEM). Based on their findings, along with the groove angle, the mechanical properties increase, yet the homogeneity of the equivalent plastic strain distribution reduces. All previously reported die designs were corresponded to parallel grooves or parallel studs. However, based on the effect of die profile using FEM, few reports have been concerned on relation between strain distribution and residual stress distribution in details.

The current research focuses mainly on studying the variables contributing with the distribution of equivalent total strain, the distribution of equivalent plastic strain, residual stress, stress concentration, grain size and the dislocation density of copper sheets using modified-constrained studded pressing (modified-CSP). Firstly, in the first step of the process, a three-dimensional elastic-plastic model was established for analysis of crack initiation, crack growth, and calculating the residual stress utilizing the FEM. The X-ray diffraction (XRD) method was used to study the residual stress and the microstructural evolution including grain size and dislocation density. The non-destructive penetration test (PT) was also used to examine the areas as well as the growth steps of the surface cracks.

2. Materials and methods

99.99% pure copper sheets with dimensions of 120 × 50 × 2 mm were used in this study as raw material. To eliminate the residual stress and improve the homogeneity of microstructure, the raw materials were annealed at 650 °C for 3 h in a resistance heated furnace. After annealing, hydrochloric acid was used to deoxidize and clean the copper sheets.

Corrugation and flattening of the sheets were carried out by the modified-CSP dies on a 20-ton hydraulic press machine at room temperature with a constant speed of 0.1 mm s⁻¹. Figure 1 shows the modified-CSP set up dies and its samples. The corrugation set-up dies, the flattening set-up dies, and the annealed, corrugated, and straightened sheet samples used for the first pass of the process are shown in figures 1(a), (b), and (c), respectively. Each corrugation and flattening step of the sheets was considered as a separate pass. Also, the modified-CSP passes were performed up to tenth passes.

X-ray diffraction (XRD) was used to analyze the characteristics of the microstructure of the modified-CSPed copper sheet samples. A PW1730 X-ray diffractometer which uses a copper lamp to produce X-rays with a wavelength of 1.54 Å and an accuracy of 0.05 degrees (2θ) was further used. Full-width at half maximum (FWHM), peak position (2θ), and peak area (A) were calculated from the XRD patterns. The instrumental effect for line broadening was calculated from Warren’s method. Finally, the XRD results were used to approximate
grains size and dislocations density. The PT was also used to detect surface micro-cracks and crack propagation stages. From each pass, ten specimens were prepared for PT. First, for the PT test, the surface of the processed sheets was cleaned from surface contaminants. Spraying fluorescent penetrant on the surfaces took about 1 h (dwell time) for the penetrant to penetrate the discontinuities. As the dwell time ended, the excess penetrant was removed from the surface. In the last step, the developer material was applied after cleaning the excess penetrant. Photography was utilized to observe the places of accumulated developer material and determine the location of crack initiation. To observe the microstructural evolution of the samples, the optical microscope and the field emission scanning electron microscope (FE-SEM) model TESCAN MIRA3 were respectively used before and after the process. The metallographic samples were first prepared from the middle of the sheet samples with dimensions of $2 \times 1 \times 1$ mm and were then mounted. After initial polishing with sandpaper No. 200, 2000, 1000, and 2500, the samples were polished at the speed of 500 rpm and were then washed with alcohol. Keller solution with a combination of 2 ml HF, 4 ml HCl, 20 ml HNO3, and 175 ml distilled water was also applied for etching the samples. For determination of the residual stresses induced on the surface of samples used the XRD-$\sin^2 \Psi$ method. Set-X instrument with Cr-K$_\alpha$ radiation running at 30 mA and 40 kV with the wavelength equal to 2.2 Å used as anode. This method mainly bases on the measurement of the shift in a diffraction peak position recorded for various $\Psi$ angles \cite{13}. The samples in 3 conditions (as annealed, first pass, and tenth pass) have been opted for evaluating the amount of residual stresses by XRD-$\sin^2 \Psi$ method. Accordingly, 10 $\Psi$ angles between a $2\theta$ angles from 126° to 129° were used for stress analysis. Face center cubic lattice and crystallographic planes, $\{200\}$ ($2\theta = 127.2^\circ$, step size equal to 0.03° and the probe depth equal to 20 $\mu$m) were used. It should be noted that, the residual stresses can be calculated with the use of elastic constants of the studied material: Poisson coefficient ($n = 0.3$), Young modulus ($E = 110$ GPa), and the anisotropic factor ($\Delta_{KX} = 1$). The Croto computer software was used for analysis of the results. The portable diffractometer used for X-ray diffraction was the Proto residual stress analyser. The Gaussian curve fitting was done for analysis of the peak value of Bragg angle at each measurement location.

3. Finite element simulation procedure

FEM was adopted to study the distribution of equivalent total strain, the distribution equivalent plastic strain and residual stress analysis in the modified-CSP process. The DEFORM-3D was also used for simulation. The bottom and top dies were considered rigid. The sheet was made of copper with elastic-plastic behavior. The Johnson–Cook equation was considered as a material model to express the mechanical behavior of the copper samples. The Modified-CSP process were simulated using a particular isotropic hardening model proposed by Johnson–Cook. The Johnson–Cook model is a phenomenological model for materials subjected to large strains,
high strain rates and high temperatures. The flow stress of this particular type of Mises plasticity model is expressed as [14]:

$$
\sigma = (A + B\varepsilon^n) \left(1 + C \ln \left(\frac{\varepsilon}{\varepsilon_0}\right) \right) \\
\times \left[1 - \left(\frac{T - T_r}{T_m - T_r}\right)^m\right] \tag{1}
$$

where $A$, $B$, $n$, $C$, and $m$ are the material constants; $\varepsilon$ is the equivalent plastic strain, $\varepsilon$ and $\varepsilon_0$ are the plastic strain rate and the reference plastic strain rate (i.e., which corresponds to the quasi-static test), respectively. $T_r$ is a reference (or transition) temperature (i.e., the room temperature in a quasi-static test), and $T_m$ is the melting temperature. $A$ is defined as the initial static yield stress. terms of $(A + B \varepsilon^n)$, $\left(1 + C \ln \left(\frac{\varepsilon}{\varepsilon_0}\right) \right)$, and

$$
\left[1 - \left(\frac{T - T_r}{T_m - T_r}\right)^m\right] \text{ in equation (1) describe the effects of strain hardening, strain rate hardening and temperature softening of the metallic materials, respectively. The expression of a first set bracket in equation (1) represents the stress as a function of strain. Where, parameters } A, B \text{ and } n \text{ can be determined using static tensile test. The expression of second bracket set represents an effect of strain rate hardening with the parameter } C. \text{ The last set of expression represents the temperature sensitivity. Table 1 shows the constant of the material model.}

The physical and mechanical properties of copper sample are shown in table 1.

For the purpose of 3D simulation, the specimens were meshed by the tetrahedral element. The coefficient of friction between the sheet and the dies was Coulomb type and 0.1. The top die speed was considered 0.1 mm s$^{-1}$. In order to investigate the convergence and independence of the solutions from the mesh, the simulation was conducted with different numbers of meshes. So, simulations were performed for 20,000, 25,000, 40,000, 55,000, 50,000 and 55,000 elements. The equivalent plastic strain values for the 50,000 and 55,000 meshes were almost identical.

The simulation of first corrugation process was performed in two stages: loading and unloading. First, like the real pressing operation, the die was moved down and up. In the second step, the die was returned up and loaded at the same speed as before. Also, the distribution of equivalent total strain, the distribution of equivalent plastic strain and residual stresses was also investigated. Figure 2 shows the model of modified-CSP process from the DEFORM-3D. In accordance with figures 2(a) and (c), a copper sheet was located respectively between a pair of the corrugation and flattening dies for simulation. Also, figures 2(b) and (d), illustrate the modified-CSPed sheet after corrugation and flattening steps, respectively.

Figures 3(a) and (b) show the distribution of equivalent total strain and equivalent plastic strain in loading and unloading steps, respectively. According to figure 3(a), at the end of the loading stage, the maximum equivalent total strain of Von Mises is 1.5, and the pressing force is at its maximum value. However, according to figure 3(b), after unloading, the maximum equivalent plastic strain of Von Mises has been reduced to 1.2. All studies are based on the behavior of the material after loading. In general, according to the finite element simulation results, the equivalent plastic strain distribution is relatively uniform. Moreover, the amount of strain applied in the modified-CSP process is 1.2.

### 4. Results and discussion

#### 4.1. Residual stress analysis

Figure 4 depicts the residual stress distribution occurring after unloading in the corrugation and flattening steps of modified-CSP resulting from the finite element simulation. According to figure 4, lines A and B were considered for analysis of normal residual stress distribution ($\sigma_n$). In figure 5, the normal residual stress distribution ($\sigma_n$) in the first corrugation and flattening steps along line A is showed. Likewise, in figure 5(a), the residual stresses along line A are tensile and compressive. Thus, the residual stress in corrugation step ranged from +20 MPa to −26 MPa, whereas in figure 5(b), the normal residual stress distribution ($\sigma_n$) along line A for the flattening step changed completely into compressive state. The maximum compressive residual stress along line A for the flattening step is −3.5 MPa.

| Material | $A$ (MPa) | $B$ (MPa) | $C$ | $n$ | $m$ | Density (kg m$^{-3}$) | Young modulus (GPa) | Poisson’s ratio | Reference |
|----------|-----------|-----------|-----|-----|-----|-----------------------|--------------------|-----------------|----------|
| Copper   | 90        | 292       | 0.025 | 0.31 | 1.09 | 8.93                  | 125000             | 0.34            | [14]     |
As shown in figure 6(a), the residual stress in corrugation step along line B is completely tensile. The maximum tensile residual stress along line B in the corrugation step is $+32$ MPa. However, as shown in figure 6(b), the residual stress along line B for the flattening step varies from tensile to compressive state. Since the cracks are able to initiate and grow in the tensile residual stress fields, the compressive residual stresses can delay crack initiation and growth [15]. Therefore, according to results in figures 5(b) and 6(b), after the flattening step, the compressive residual stress acts as an important factor for preventing cracks initiation and growth.

### 4.2. Residual stress measurement

Residual stress for the as annealed, first pass, and tenth pass of modified-CSPed samples were measured by the X-ray diffraction (XRD) method. Figures 7(a), (b), and (c) represent the X-ray diffraction patterns for the as annealed, first pass, and tenth pass of modified-CSPed samples, respectively. In comparison with the as annealed...
sample, the width of the peaks increases along with an increase in the amount of plastic strain in the first pass and tenth pass. A main reason for such increase in the width of XRD peaks is the grain refinement and strain in the crystal lattice [16]. As shown in the figure 7, the crystalline planes for three of the samples is (2 0 0).

According to the experimental results, the residual stress in the as annealed copper sample, the first pass, and the tenth pass of the modified-CSP process are +128.2 MPa, +80.4 MPa, and −62.5 MPa, respectively. By applying the amount of plastic strain at tenth passes, the residual stress changed from tensile to compressive. As mentioned earlier, the compressive residual stress on the samples can delay the initiation and growth of cracks [15].

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**Figure 4.** Residual stress distribution after (a) corrugation step, (b) flattening step.

**Figure 5.** Normal residual stress ($\sigma_x$) distribution along line A in (a) corrugation and (b) flattening steps.

**Figure 6.** Normal residual stress ($\sigma_x$) distribution along line B in (a) corrugation and (b) flattening steps.
4.3. Analysis of crack initiation and growth and its correlation with results of FEM analysis

The stress concentrations sites for corrugated modified-CSPed sheet sample are represented in figure 8 where the stress concentration which occurs in surfaces of each valley is the collision of 4 studs. Lack of an interface region (sharp edges) in the modified-CSP dies makes more pressing cycles with no early crack formation in sample than in the CGP, and then in CSP dies. Later, following the modified-CSP passes, it causes strain accumulation in the sheet around the valley between the studs, that is, one of the most prone sites for cracks initiation and growth.

Figure 9, illustrates the sites and stages of crack initiation and growth versus pass numbers for the modified-CSPed samples, which were detected by the PT method. According to figure 9(a), in the first to fourth pass samples, the PT observations showed that cracks were not created. In figure 9(b), the crack was observed for the fifth pass sample. According to figure 8 and PT observations, the area between the 4 studs was an optimal site for crack initiation which led to its growth. Interestingly, each pass includes two steps: corrugation and flattening. In the latter, the reverse loading on the corrugated sample causes strain accumulation, mainly due to bending in the first points of collision of studs of upper and lower corrugated dies. According shown in figure 9(c), through
continuing the process, the cracks were propagated from the sixth to ninth passes. Finally, in figure 9(d), the cracks meet in the tenth pass, and the results in the eleventh pass showed that the sheet was rupture.

4.4. Microstructure observations

Figure 10 shows an optical microscope image of the annealed copper sheet. The Heyn method was also used to measure the grain size. According to figure 10, the grain size of the sheet with an average of ~35 micrometers is heterogeneous. Figure 11 shows an optical microscope image of the first pass specimens for the CGP, CSP, and modified-CSP methods. According to figure 11, the average grain sizes for the first pass of CGP, CSP, and modified-CSP methods were about 28, 27, and 14 micrometers, respectively. As shown in figure 11, the grains size reduced by ~ 50% for the modified-CSP method in comparison with the CGP and CSP methods, mainly because more strain is applied. The shape of the grains in the CGP has an elongated morphology. The grain elongation in the CGP sheets was also reported by Torkestani and Dashtbayazi [3]. The reason for the elongation of the grains in the CGP is the application of strain in the direction of the grooves. More homogeneous strain distribution in the CSP and the modified-CSP specimens was caused by the presence of bilateral grooves and applied strain was more than the CGP, so it was reported that the grains were coaxial and finer. The grain size refinement in the first pass was recorded in all three methods so that the grain size was significantly reduced due to this severe plastic deformation.

Figure 12 shows SEM and FE-SEM micrographs which were taken from the tenth pass specimen of the modified-CSP process. Twin bands are seen in the figure 12(a). Here, two basics of plastic deformation mechanism are slip and twinning. In the initial process passes, the plastic deformation is resulted by sliding. Meyers et al [17]
Figure 10. Optical microscope image of annealed copper sheet.

Figure 11. Optical microscope images of the first pass samples for (a) CGP, (b) CSP and (c) modified-CSP.

Figure 12. (a) SEM micrograph of tenth pass for modified-CSP process and (b) FE-SEM micrograph for tenth pass of modified-CSP process.
reported the onset of mechanical twinning in metals with low stacking fault energy (SFE) like pure copper requires a critical dislocation density. As the minimum dislocation density and critical shear stress are created, the mechanical twinning mechanism is activated so that the mechanical twinning occurs at high strains. It can also be a critical dislocation density. As the minimum dislocation density and critical shear stress are created, the 

\[\text{dislocation density} \times \text{critical shear stress} = \text{critical dislocation density} \]

shows XRD diffraction pattern for the annealed, the 

for high strength. Although strength and ductility appear to be mutually incompatible in most cases, Höppel [18] reported the onset of mechanical twinning in metals with low stacking fault energy (SFE) like pure copper requires a critical dislocation density. As the minimum dislocation density and critical shear stress are created, the mechanical twinning mechanism is activated so that the mechanical twinning occurs at high strains. It can also be inferred from figure 12(b) that the grain size of the specimen which was produced in the tenth pass was extremely fine. After the tenth pass of the modified-CSP process, copper with an average crystallite size of 76 nm was obtained. According to figure 12(b), the FE-SEM micrograph illustrates coarse grains, nanocrystalline and UFG’s. One micrometer grains in a matrix of UFG and nanocrystalline grains are called bimodal grain size. Vinogradov [18] was claimed that the coarse grains led to high ductility, while the nanocrystalline and UFG’s were responsible for high strength. Although strength and ductility appear to be mutually incompatible in most cases, Höppel et al [19] have shown that strength and ductility can increase simultaneously.

The grain size evolution and the dislocation density were calculated by XRD pattern analysis. Figure 13 shows XRD diffraction pattern for the annealed, the first, and the tenth passes of the copper sheet samples. According to figure 13, the peak width increased along with an increase in plastic strain for the first and tenth passes compared to the annealed sample. A main reason for increasing the width of XRD peaks is the grain refinement and strain in the crystal lattice. The highest intensity of XRD is due to the diffraction of crystalline planes (1 1 1), (0 0 2), and (0 2 2) at angles of 43.53°, 50.65°, and 74.30° for as annealed sample. Also, the highest intensity of XRD is due to the diffraction of crystalline planes (1 1 1), (0 0 2), and (0 2 2) at angles of 43.70°, 50.84°, and 74.50° for the first pass sample. The highest intensity of XRD is due to the diffraction of crystalline planes (1 1 1), (0 0 2), and (0 2 2) at angles of 43.87°, 50.99°, and 74.59° for the tenth pass sample.

Due to the application of severe plastic deformation, both the peak broadening behavior and the shift peak behavior occur. This result was also reported by other researchers [20, 21]. In fact, the peak broadening behavior is related to grain refinement and dislocation density [22]. As shown in figure 13 and table 3, the amount of strain applied during the severe plastic deformation process increases along with the density of dislocation. Increased density of dislocation occurring during the first and tenth passes of the Modified-CSP in turn creates high angle grain boundaries, and leads to grain size reduction occurs indicating a peak broadening behavior in grain size at XRD peaks. The relationship between the peak broadening behavior and the grain refinement is proportional.

According to figure 13, the difference between XRD peaks is insignificant. Table 2 shows the angles for each crystalline plane in each sample. Also, according to figure 13, all the peaks are shifted only to the right. Moreover,
according to figure 3, the geometric design of the modified-CSP dies makes the equivalent plastic strain distribution more uniform. The presence of this distribution caused diffraction line shifts. The peak shift under uniform strain has also been reported by other researchers [21].

The grain size for different modified-CSPed passes was calculated by the Williamson-Hall method. The dislocation density and the grain size relationships were examined. The dislocation density ($\rho$) was calculated from [23]:

$$\rho = \left(\rho_D \rho_g\right)^{0.5}$$  \hspace{1cm} (2)

$$\rho_g = k(\varepsilon^2)/b^2$$  \hspace{1cm} (3)

$$\rho_D = (3/D^2)$$  \hspace{1cm} (4)

where $\rho_D$ and $\rho_g$ are dislocation due to domain size and dislocation due to lattice strain, respectively, $k$ is material constant, $D$ is size of grains, $\varepsilon$ is lattice strain, and $b$ is Burgers vector. Table 3 depicts the grain sizes and dislocations density for different modified-CSPed samples so that when the number of passes increases, the grains size greatly decreases after the first pass (from $\sim 35 \mu m$ to $\sim 14 \mu m$) for an annealed sample. By continuing the process up to the tenth pass, the grains size decreases to 76 nm.

Regarding its cause, it is accepted that the plastic deformation can be related to the evolution of dislocations and the interaction of dislocations. There are several models suggested which deal with generation and annihilation of dislocations in plastic deformation of polycrystalline metals [24, 25]. Also, the relationship between the microscopic plastic deformation mechanism and the notion of dislocation is generally accepted. As shown in table 3, the annealed sample has low dislocation density, whereas the modified-CSPed samples can increase dislocation density. In the first pass, due to the SPD, the number of dislocations increases sharply. As the process continues, during the first to fifth passes, increase in dislocation increases is followed by decrease in grain size, and finally, high-angle grain boundaries (HAGBs) are formed. From the fifth pass to the tenth pass, due to the simultaneous increase of the dislocation and their collisions with each other, the force required to move the dislocation also increases. That is, an increase occurs in the strength of the sample. Along with this evolution, the grain size decreases greatly and the bimodal grains are formed. The bimodal grain size was proposed for coexisting of high ductility and high strength by Valiev and Langdon [26]. This has been suggested by Krishnaiah et al [27] for the UFG metals in a way that ductility increases at large strains in the SPD processes are caused by increases in HAGBs fraction.

5. Conclusions

Residual stress and microstructural evolution of copper sheets which were produced by modified-CSP have been investigated excrementally and numerically. A finite element model was established to analyze the residual stress and crack initiation. The residual stress of the samples was investigated using X-ray diffraction method. Due to significance of surface micro-cracks, both crack initiation and propagation were investigated by the penetration test. The most important results are as follows:

1. In general, the results of finite element simulation indicate that distribution of equivalent plastic strain is relatively uniform. In addition, the quantity of equivalent plastic strain applied in the modified CSP process is 1.2.

2. Due to the strain localization in collision of 4 studs, the stress concentration in this place experiences the maximum value. By repeating the process and taking into account the strain accumulated in this place, these surfaces are among the most prone places for cracks initiation and propagation. The results of the non-destructive PT are in good agreement with the simulation results.

3. According to X-ray diffraction patterns, the peak broadening behavior and the shift peak behavior indicate the grain refinement and uniform distribution of the equivalent plastic strain in the modified-CSP sample.

| Samples         | $\rho$ (1 nm$^{-2}$) | Grain size (nm) |
|-----------------|----------------------|-----------------|
| As annealed     | 1.003                | 35000           |
| First pass      | 1.841                | 14000           |
| 10th pass       | 2.765                | 76              |

Table 3. The values of the grain size and dislocation density in as annealed, the first and tenth pass of modified-CSP method.
4. In the modified-CSP, for the residual stresses diagram, the width of the peaks increased along with increase in the amount of plastic strain in the first and tenth passes compared to the annealed sample.

5. Regarding the redistribution of residual stress in different locations, it can be determined that increasing the number of passes delays the onset and growth of cracks, mainly due to changes occurring in the sign of residual stress from tensile to compressive.

6. By increasing pass numbers, the density of the dislocations increases so significantly while grain size decreased sharply.

Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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