

Manufacturing a Curved Profile with Fine Grains and High Strength by Differential Velocity Sideways Extrusion

Wenbin Zhou^a, Junquan Yu^{a,*}, Jianguo Lin^a, Trevor A. Dean^b

^a*Department of Mechanical Engineering, Imperial College London, London SW7 2AZ, UK*

^b*Department of Mechanical Engineering, University of Birmingham, Birmingham B15 2TT, UK*

Abstract

For structural and aerodynamic reasons, curved profiles are widely used in the transport industry for manufacturing lightweight structures. In the present work, a curved AA1050 bar with fine grains and high strength was manufactured by a novel forming technique, differential velocity sideways extrusion (DVSE). The evolution of grain structure and micro-texture during DVSE and the mechanical properties of the formed bar were studied, and the grain refinement mechanism was revealed. Due to the severe strains arising in the process, (greater than that for conventional one pass equal channel angular extrusion), significant grain refinement in the curved bar (grain size $\sim 3 \mu\text{m}$) was achieved from the original billet (grain size $\sim 357 \mu\text{m}$) in one extrusion operation. Coarse band-like structures containing subgrains with low angle boundaries in the shearing zone gradually transformed into fine shear band-like structures containing equiaxed (sub)grains with a mixture of low and high angle boundaries. The fine shear band-like structures inclined approximately along the shear intersection planes. Severe plastic deformation induced a high dislocation density that initiated subgrain walls with low angle boundaries, which gradually transformed into grain boundaries with high misorientation, indicating that refinement of AA1050 grains in the DVSE process is due mainly to continuous dynamic recrystallization (cDRX). Due to the appearance of greater effective strain on the inner bend of the extruded bar, the grain

* Corresponding author: junquan.yu@imperial.ac.uk.

refinement degree and high angle boundary fraction of the material on the inner bend are slightly greater than those of the material on the outside. DVSE resulted in a weak C-type shear-texture component which can be determined by a proper rotation of the negative simple shear texture. Compared with the billet, significant increase of the hardness, yield strength and ultimate tensile strength by 134.8%, 354.0% and 116.8% respectively was achieved in the formed curved bar, although the elongation to fracture was decreased by 60.0%.

Keywords: Differential velocity sideways extrusion (DVSE); Curved profile; Aluminium alloy 1050; Severe plastic deformation (SPD); Grain refinement; Continuous dynamic recrystallization (cDRX)

1. Introduction

Due to their potential for light weight, high stiffness and ease for incorporating into aerodynamic forms, curved profiles with either solid or hollow cross-sections are extensively employed as constructional elements in transportation structures. Curved profiles can be utilised to construct lightweight structures with few joining and welding efforts compared with straight profiles, thus showing improved space saving and aerodynamic properties [1]. Curved profiles are also widely used in civil engineering and architecture such as the door/window frames and building facades due to their unique aesthetics and structural design reasons [2].

Various forming techniques, such as the stretch bending, rotary draw bending, press bending, and roll bending, have been developed for manufacturing curved profiles, and much research has been conducted to deeply understand and improve these operations [3-8]. Detailed reviews were made by Vollertsen et al. [9] on curved profile forming techniques and Yang et al. [10] specifically on tube (a type of profile with a hollow tubular cross section) bending forming technologies. Most of current curved profile forming methods include more than one process, normally starting with conventional extrusion or rolling to get semi-

finished a straight profile followed by a bending operation which is an added manufacturing cost and reduces productivity. Also the bending process is associated with problems such as cross-section distortion, wrinkling and springback. Although some novel stress-superposed bending methods have been proposed to alleviate the bending defects [11-13], and extrusion-bending integrated methods have been developed to minimise the forming procedures and improve production efficiency [14-16], no obvious grain refinement or improved mechanical property is actually realised in current curved profile forming processes.

Grain size is the most important and dominant structural parameter in polycrystalline materials. In low-temperature regimes, i.e. below $\sim 0.5T_m$ where T_m is the absolute melting temperature, a small grain size is advantageous since it results in a significantly increased strength. In high-temperature regimes, a small grain size provides the possibility of superplastic forming at rapid rates which may be employed in industrial forming operations [17]. To improve both the forming capability and mechanical properties, thus giving promise for structural applications [18], grain refinement in bulk metallic materials down to the sub-micrometer or even nanometer range [19] have been widely studied by severe plastic deformation (SPD) processing technology. SPD can be used to produce ultra-fine grained (UFG) materials with finer grains than those obtained in conventional thermomechanical processing [17]. Various SPD techniques have been developed such as equal channel angular pressing/extrusion (ECAP/E) [20], high-pressure torsion (HPT) [21], accumulative roll bonding (ARB) [22]. Some modified ECAP processes were proposed for the purpose of achieving greater efficiency in grain refinement, such as 2-turn ECAP [23] which adds one more turn to the channel, and non-equal channel angular pressing (N-ECAP) [24] in which the outgoing channel has a reduced diameter. To date, the commercialisation of UFG products/structures in industry using SPD processed materials have been concentrated on

niche applications where material cost is not a major concern, such as for medical implants [25], defense applications [26], and aerospace components [27], etc.

Currently, the ultra-fine grained material used in these components is made prior to the shape forming operation. Although SPD can improve the strength of most materials by refining grain size, formability at room temperature, exemplified by aluminium alloys, is generally reduced [28]. It is to the contrary for magnesium, where yield strength reduction and the drastic ductility enhancement in AZ31 [29] and AZ61 alloy [30] at room temperature tension after ECAP were found due to the texture softening is dominant over the strengthening caused by grain refinement. In addition, the fine grain obtained by SPD detracts from making parts by plastic deformation, especially by hot processing. Lee et al. [31] fabricated a forged impeller using grain refined magnesium alloy after ECAP, it was found that the fine grain size and microhardness obtained by ECAP were affected by processing temperature during forging. More importantly, the factor most limiting industrial use of fine grained material is the cost in producing it with SPD and the production efficiency.

A novel process integrating forming, bending and grain refining is promising to improve the production efficiency and mechanical properties. The authors have developed differential velocity sideways extrusion (DVSE) process to directly manufacture curved profiles from conventional billet material in one operation [32,33]. It has been proven that profiles of predetermined accurate curvature and undistorted cross-section can be produced. An initial investigation indicated that severe plastic strains arise in the DVSE process, so it is promising to refine grains during the forming process and enhance mechanical properties of profiles. In this paper, a curved AA1050 bar with fine grains and high strength was manufactured by the DVSE process at room temperature. Through a thorough microstructural and mechanical examination of the formed curved bar, the evolution of grain structure and micro-texture

during DVSE and the mechanical properties of the formed bar were studied, and the grain refinement mechanism was revealed.

2. Experimentation and simulation

2.1. Extrusion Tools

Design and operation of the extrusion tools have been explained in detail in, Zhou et al. [33] and for illustration, only a schematic is given in **Fig. 1a**. The tool-set, for tests described here, consisted essentially of a, vertically oriented, billet containing, circular container of 25.6 mm bore with circular exit die orifice, 20 mm diameter, located in its side, equidistant from the ends. Punches at each end of the container, which can be moved independently at different velocities, are used to compress the billet, which, as the axis of the orifice is normal to its axis, extrudes sideways, which if the punch velocities are different from each other, to form a curved profile. The extrusion tool was split in the common plane of the container and orifice axes, so that the halves could be separated to extract the extruded shape.

2.2. Experimental method for DVSE

A cylindrical billet of annealed (450 °C, 1 h) AA1050, 25.4 mm in diameter and 130 mm in length, was used for the test, undertaken at room temperature, in which the billet was lubricated with graphite-based Omega-35 grease. The extrusion tool-set was located in an Instron materials testing press. To extrude a profile, the billet was placed in the container and the press operated to drive in the extrusion punches. At the end of deformation, the halves of the tool were separated to extract the formed shape.

According to the previous study [33], in the process of DVSE, curvature of curved profiles can be controlled by adjusting the ratio of extrusion velocities of two opposing extrusion punches and the extrusion ratio. Lower velocity ratio and greater extrusion ratio result in greater curvature. The specific correspondence between curvature and velocity ratio and

extrusion ratio can be obtained using the analytical upper-bound-based model previously proposed by the authors or finite element analysis [34]. Here, to manufacture a curved profile, velocities of the upper and lower punches were typically set as 1.0 mm/s and 0.5 mm/s, respectively. The curved extruded profile is shown in **Fig. 1b**.

2.3. Process modelling and locations for microstructure characterisation

To analyse material flow lines and quantify the strains induced by DVSE, the DVSE process was simulated using the finite element software Deform-3D. A friction coefficient of 0.16 was used in the finite element model, which was determined by comparing simulated and experimental results of curvature and extrusion force under a series of friction coefficients [34]. The constitutive equation used for the model was $\sigma = 145.5\varepsilon^{0.296}$ [34]. Simulated shapes with flow lines and induced strains are shown in **Figs. 1c** and **d**, respectively. Generally, shape is simulated accurately, although a small difference exists between real and simulated curvature at emergence of extrusion from orifice. **Fig. 1e** illustrates deformation model on the symmetrical diametral plane XY of the container and orifice. Here, based on previous results [34] and the simulated flow line shown in **Fig. 1c**, material in the channels of the container and orifice of exit die can be divided into two parts by a dividing line BG, and the eccentricity ratio ξ is determined as 0.55 when $v_2/v_1=1/2$. The billet is pressed in the $-Y$ and $+Y$ directions (normal direction ND) in the vertical container and flows in the $+X$ direction (extrusion direction ED) through the horizontal die orifice. The Z direction (transverse direction TD) is normal to the symmetry plane XY of the container and orifice. Deformation takes place by shear confined to two narrow near triangular zones, EPQ and HMN, around the two ideal intersection planes (which can be shown two-dimensionally by lines 'EB' and 'HB' in **Fig. 1e**) of the container and orifice [35]. This happens especially when there is a dead zone and the flow line is rounded [36]. The deformation history for a representative small square element of the material pressed through

the upper part of the die is shown in **Fig. 1e**; it comprises three successive steps: (i) shear along slip line EP, (ii) shear along α -slip lines (flow lines) inside the central fan EPQ, (iii) shear along slip line EQ.

To study the evolution of grain structure and micro-texture, the curved bar was cut normal to its axis ~10 mm from the orifice to reveal its cross-section. Three locations marked as, CS_u , CS_m , CS_l corresponding to the area near top (outer curve), centre and bottom (inner curve) were investigated. The extruded curved shape was also cut in two along the symmetry plane XY of container and orifice. Six different locations, three in the intense shearing zones in the deformed billet (S_{u1} , S_{u2} , S_l), and three in the extrudate (F_u , F_m , F_l), were studied in this plane (also called flow plane XY), as shown in **Fig. 1e**. Here, S_{u1} , S_{u2} reflect the microstructure resulting from shearing at line EP and in zone EPQ respectively, and F_u , F_m reflect the microstructure after passing shearing zone EPQ. S_l and F_l are chosen to enable comparison of microstructures of material emerging from upper and lower billet regions. The locations of these selected points are also shown in the simulated profile with strains (**Fig.1d**).

2.4. Sample preparation for microstructure characterisation and mechanical property test

Microstructures were observed using a Bruker's QUANTAX electron backscatter diffraction (EBSD) device mounted on a Hitachi S-3400N SEM, operated at a voltage of 20 kV. Samples were prepared by conventional metallographic procedures, which include cold mounting, grinding and polishing. The cold mounting was conducted using conductive resin Demotec 70 at room temperature. After mounting, samples were ground using 600, 800, 1200 and 4000 SiC papers. Then, an initial polishing was carried out using Metprep® Alpha cloth, 1 μ m diamond polishing solution and water-based lubricant, and a further polishing was performed using Metprep® Alpha cloth and 0.04 μ m colloidal silica solution. Finally,

samples were ion polished for EBSD characterisation. Hardness tests were conducted at the above selected locations using a Zwick Roell machine with an applied force of 10 kgf (HV10) and dwell time of 10 s. Room temperature tensile tests were carried out on a Deben MICROTTEST tensile stage at a speed of 0.4 mm/min using specimens with gauge sections of 2 mm×3 mm×5 mm, machined from the original billet and curved bar along its axis.

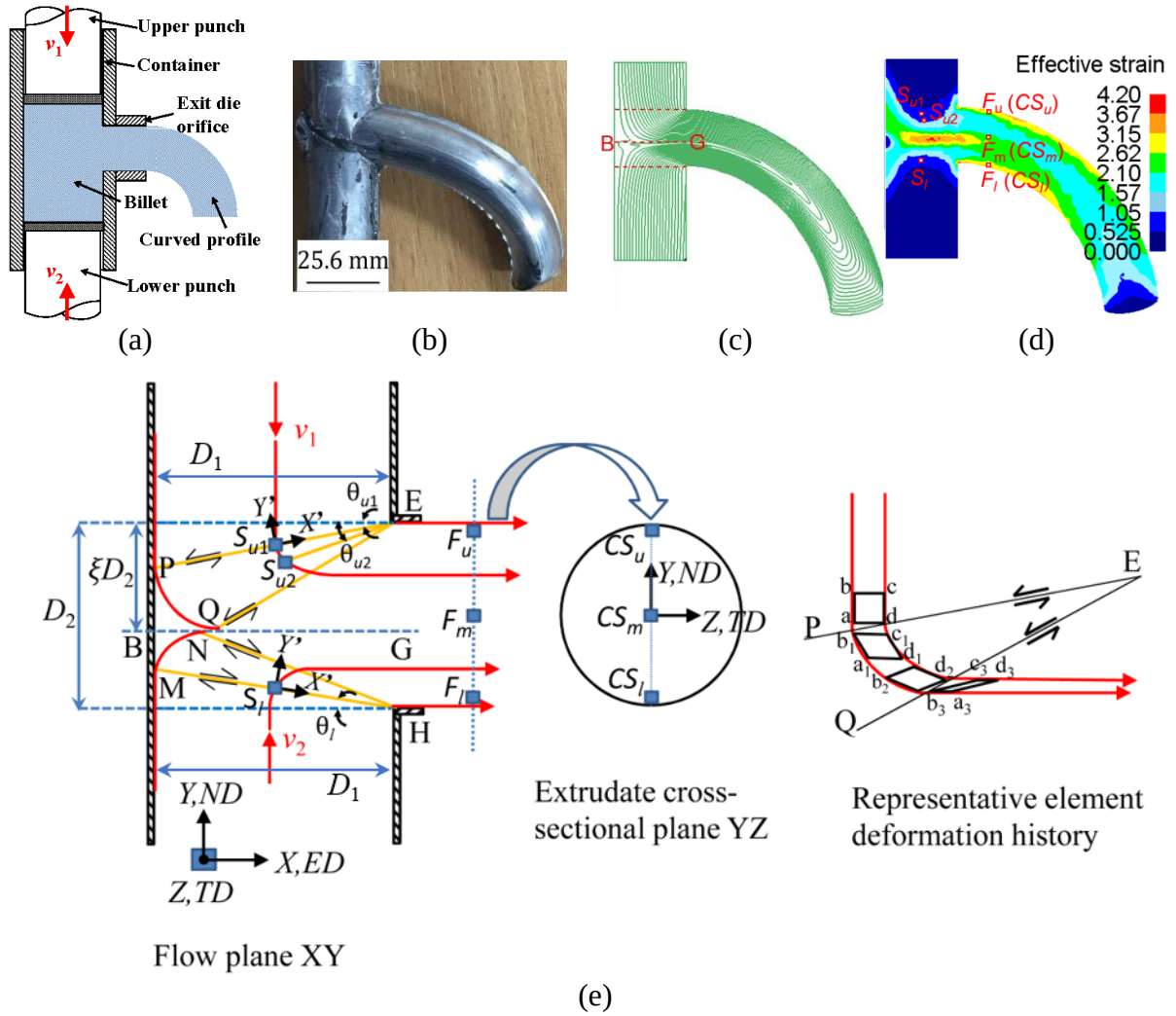


Fig. 1. (a) Schematic of the differential velocity sideways extrusion (DVSE) process, (b) the formed curved bar, (c) the simulated curved bar and flow line, (d) distribution of effective strain, (e) two-dimensional deformation model based on the flow line and selected locations for EBSD characterisation.

3. Results and discussion

3.1. Grain structure

Microstructure in the cross-sectional plane of the original billet are shown in **Fig. 2a**. The white line indicates the low angle boundaries (LABs, misorientation angle $2^\circ \leq \theta < 15^\circ$) and the black line indicates the high angle boundaries (HABs, misorientation angle $\theta \geq 15^\circ$). The grains range in size from $100\ \mu\text{m}$ to $700\ \mu\text{m}$, and the average size is estimated to be $356.79\ \mu\text{m}$. These results indicate that the grains of billet material are large and most of them have HABs. **Figs. 2b-d** show microstructures at locations CS_u , CS_m and CS_l in the cross-sectional plane YZ of the formed curved bar. It can be seen that, compared with the grains shown in **Fig. 2a**, grain size is reduced significantly and grains are more equiaxed and homogenous. It is also worth noting that the grains shown in **Figs. 2b-d** have many LABs.

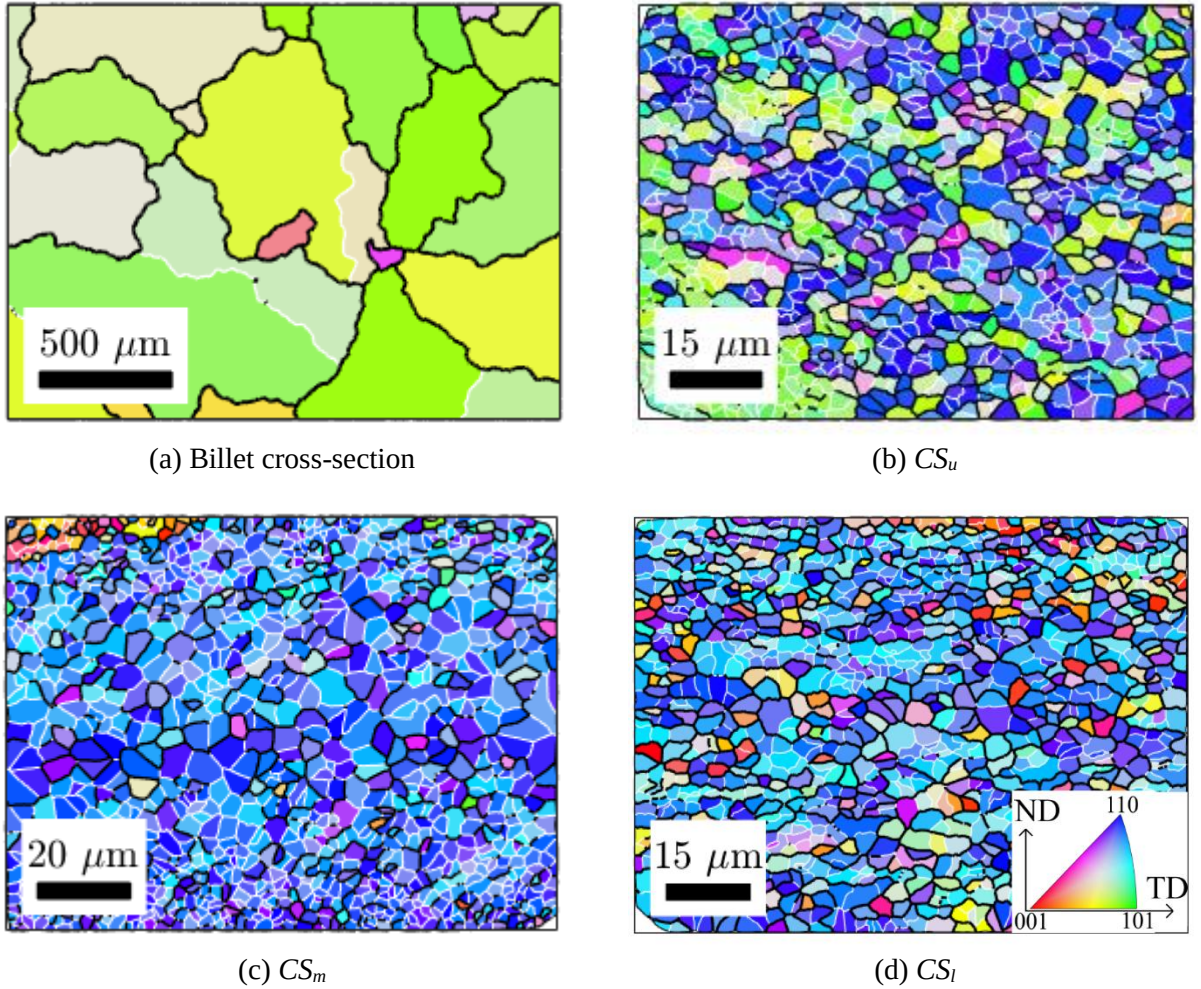


Fig. 2. Microstructure of the cross-sectional plane of (a) billet, (b) CS_u of the formed curved bar, (c) CS_m of the formed curved bar, (d) CS_l of the formed curved bar.

The specific grain size distribution is shown in **Fig. 3**, from which the average grain sizes of cross-section locations CS_u , CS_m and CS_l are estimated to be $\sim 2.84 \mu\text{m}$ on the outside, $\sim 3.32 \mu\text{m}$ near the centre and $\sim 2.69 \mu\text{m}$ near the inside of the curve of the bar, respectively. Slight differences in grain size from CS_u to CS_l indicates that the grains of the cross-section of the formed profile are almost homogeneous. Surprisingly, the average grain size of the curved profile is less than 1% of average grain size of the billet. The grain refinement for AA1050 after one pass, using ECAP with 90° channel angle, a popular grain refining extrusion process, has been reported to be 5.3% (from $600 \mu\text{m}$ down to $32 \mu\text{m}$) by El-Danaf [37], and 6.3% (from $45 \mu\text{m}$ down to $2.87 \mu\text{m}$) by Qiao et al. [38]. It should be noted that grain boundaries in these reports are defined by misorientation angles greater than 2° , which is the same as for this work. Obviously, DVSE is a very effective technique for refining grains.

In general, grain size decreases with the increase of effective strain during cold SPD [39]. Overall, the effective strain at CS_u (~ 2.75) is slightly lower than that at CS_l (~ 3.0), and the centre region CS_m has the lowest effective strain (~ 2.0), as can be seen in the simulated effective strain distribution shown in **Fig. 1d**. It can be seen that the distribution of grain size and effective strain is consistent. According to Valiev et al. [19], the effective strain in ECAP with 90° channel angle is ~ 1.0 in a single pass. In the present investigation, the lowest effective strain of DVSE is ~ 2.0 , which leads to very significant grain refinement.

Based on the previous study [34], DVSE can be reasonably regarded as two N-ECAP processes where the dividing line (BG) is shown in **Fig. 1e**. Above the dividing line, the process of extruding the material in the upper region of the container (diameter D_1) into the part with the sectional width ξD_2 can be regarded as one N-ECAP process. Correspondingly, below the dividing line, the process of extruding the material in the lower region of the container (diameter D_1) into the part with the sectional width $(1 - \xi)D_2$ can be regarded as

another N-ECAP process. When the material in the container passes through the intersection planes, its flow direction is changed by 90 degree under the action of strong shearing. As a result, SPD occurs, as can be clearly seen in the simulated results shown in **Figs. 1c-d**. The principle of material plastic deformation in DVSE is similar to that in ECAP or N-ECAP, a well-known method to achieve SPD.

Since the eccentricity ratio ξ in the present study is 0.55, i.e., $\xi D_2 > (1 - \xi)D_2$ (see **Fig. 1c and Fig. 1e**), the lower part with the sectional width $(1 - \xi)D_2$, which is formed by the material in the lower region of the container with the diameter D_1 , has more intensive deformation than the upper part with the sectional width ξD_2 . As a result, the lower part (inside bending region) of the curved profile closer to the slower moving punch has a greater effective strain.

The straining in DVSE is determined by the extrusion ratio and velocity ratio. Greater extrusion ratio will lead to a greater strain limit in general. For a given extrusion ratio, lower velocity ratio v_2/v_1 ($v_2 < v_1$) results in greater curvature and effective strain of the inner bending region, leading to greater grain refinement. The limit of the straining occurs when $v_2/v_1 = 0$, the lower part (intrados of the extrudate) will have the greatest straining as the dividing line will be closest to the lower punch (ξ will be maximum). According to the previous study [34], the strain limits for the extrusion ratio of 1.61 and 2.87 are 5.0 and 5.4, respectively.

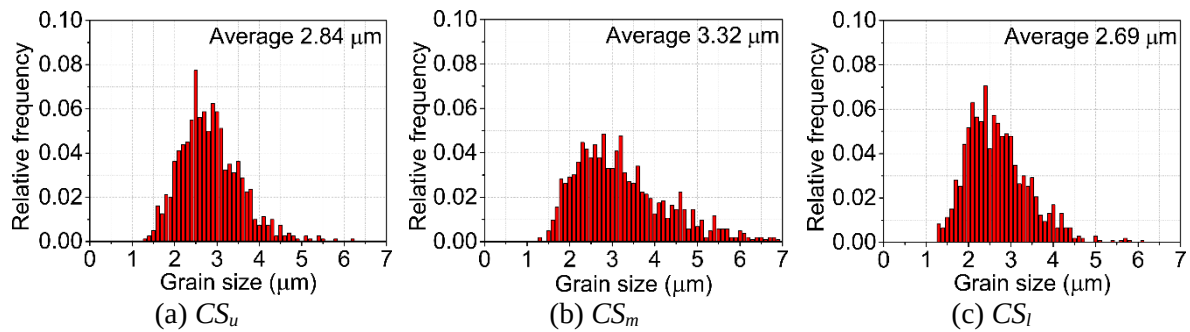


Fig. 3. Grain size distribution of the cross-sectional plane of the formed curved bar (a) CS_u , (b) CS_m , (c) CS_l .

Fig. 4 shows the specific misorientation distribution determined from the orientation imaging microscopy (OIM) data for cross-section locations CS_u , CS_m and CS_l in the formed curved bar. Fractions of LABs (f_{LABs}) for CS_u , CS_m and CS_l are 53.86%, 61.13%, and 40.16%, respectively. For comparison, the f_{LABs} on the cross-sectional plane of the billet was also calculated, and its value was 14.18%. From **Fig. 4**, it can be seen that a large number of LABs which peak at $5\sim 10^\circ$, have been generated by DVSE and compared with the billet, fractions of LABs are increased significantly. The fraction of LABs near the outer curved surface is greater than that at the inner, but smaller than that of the centre. This is due to the effective strain near the outside being slightly lower than that near the inside, and the centre having the lowest effective strain. Greater effective strain leads to greater transformation of LABs to HABs, thus a reduced fraction of LABs is achieved. The grain refinement mechanism is discussed below with reference to microstructure of the shearing zone and extrudate in the flow plane XY.

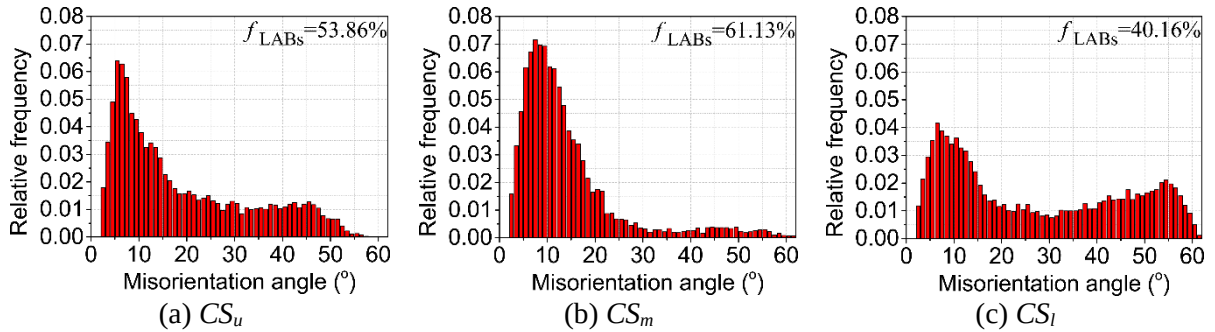


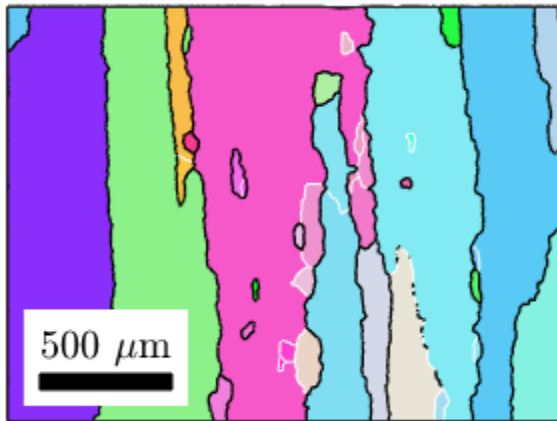
Fig. 4. Misorientation distribution of the cross-sectional plane of the formed curved bar (a) CS_u , (b) CS_m , (c) CS_l .

Fig. 5 shows EBSD grain images in the flow plane XY at different locations and **Fig. 6** gives the specific grain size distributions, from which it can be seen, average grain size gradually decreases from the S_{u1} ($\sim 27.60 \mu m$), S_l ($\sim 25.23 \mu m$), S_{u2} ($\sim 12.51 \mu m$) in the shearing zone to F_u ($\sim 2.78 \mu m$), F_m ($\sim 3.40 \mu m$), F_l ($\sim 2.49 \mu m$) in the extrudate. Before DVSE, a large number of elongated coarse grains with very few LABs exist in the billet flow

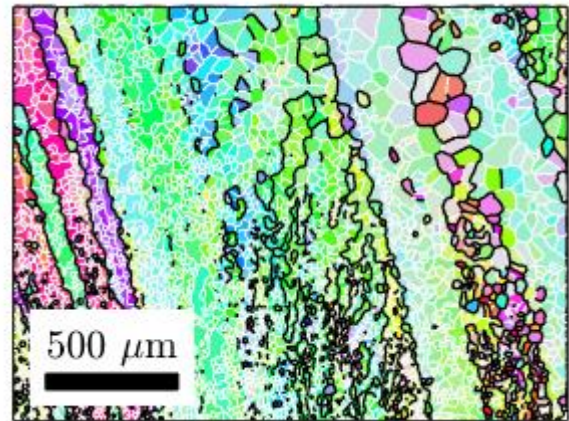
plane XY (**Fig. 5a**). After deformation, as shown in **Fig. 5b**, microstructure in the shearing zone S_{u1} is inhomogeneous, and two types of structural component are evident. One is a band-like structure consisting of the initial coarse grain boundary with HABs. This type of band-like structure tends to have rotated $\sim 15^\circ$ to the original perpendicular flow line ($-ND$ direction). The other is made up of essentially equiaxed grains with a large number of LABs (subgrains) and few HABs, within the band-like structure. Microstructure in the shearing zone S_l is shown in **Fig. 5c**. This is similar to that shown in **Fig. 5b**, but an increased rotation ($\sim 35^\circ$) to the original perpendicular flow line ($+ND$ direction) exists for the band-like structure. Additionally, the fraction of the essentially equiaxed subgrains with LABs within the band-like structure has slightly increased compared with that of **Fig. 5b**, due to the greater effective extrusion ratio resulting in greater intensive deformation. As the shearing process proceeds from S_{u1} to S_{u2} , the LABs continue to increase and band-like structure continues to rotate ($\sim 30^\circ$ to $-ND$ direction). As shown in **Fig. 5d**, there is a clear rotation around the Z axis in the shearing zone from S_{u1} to S_{u2} . **Figs. 5b-d** show that the intense shearing deformation during DVSE leads to the rotation of the band-like structure with HABs and the formation of equiaxed grains with mostly LABs (subgrains) and few HABs.

Microstructure in the flow plane XY of the curved bar is shown in **Figs. 5e-g**. Again two types of structural components are evident. One is a band-like structure consisting of elongated grains, and the size of the band-like structure has been greatly reduced compared with that of the shearing zone. The angle between elongated grains and the extrusion direction (ED) increases from $\sim 20^\circ$ (F_l) to $\sim 25^\circ$ (F_u), which is aligned approximately along the ideal shear direction (ideal intersection planes ‘EB’ and ‘HB’ in **Fig. 1e**, which are at angles of $\sim 18.5^\circ$ and $\sim 24.2^\circ$ to the $+X$ extrusion direction respectively). The other is made up of essentially equiaxed grains with a mixture of LABs and HABs within the band-like structure. Compared with that of the shearing zone, the grains are refined remarkably and the

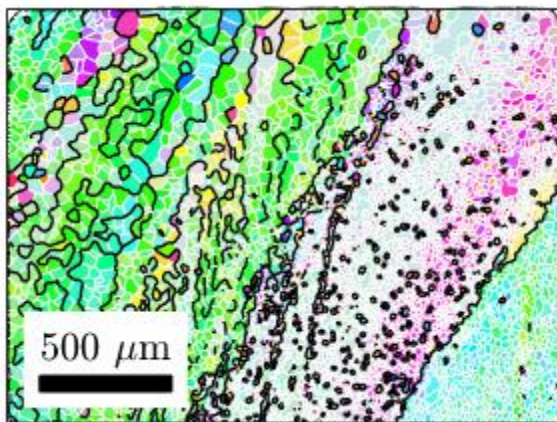
gradual appearance of colour, which represents misorientation of grains, become more apparent with increased strain, indicating that grains with higher misorientation have been generated. These results reveal that the refinement of AA1050 grains in the DVSE process is due mainly to dynamic recrystallization. That is, new grains are generated by rotation of subgrains.



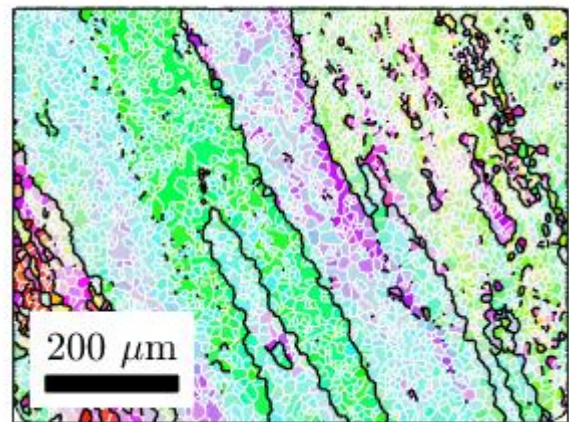
(a) Billet flow plane



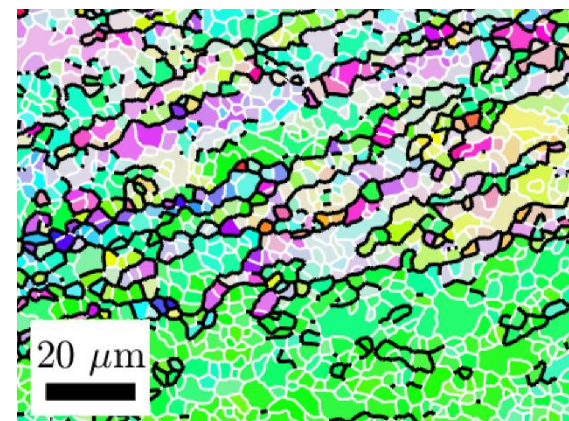
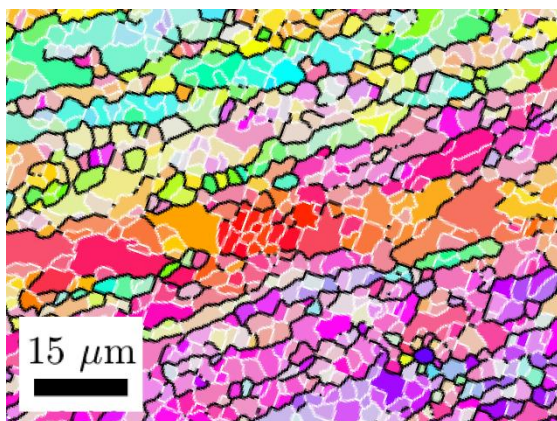
(b) S_{u1}



(c) S_l



(d) S_{u2}



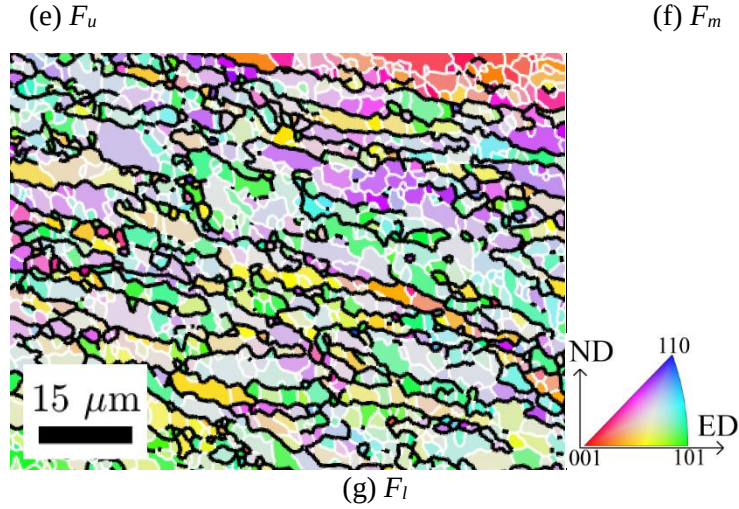


Fig. 5. Microstructure of the flow plane of (a) billet, (b) shearing zone S_{u1} , (c) shearing zone S_l , (d) shearing zone S_{u2} , (e) outer part F_u of the formed curved bar, (f) middle part F_m of the formed curved bar, (g) inner part F_l of the formed curved bar.

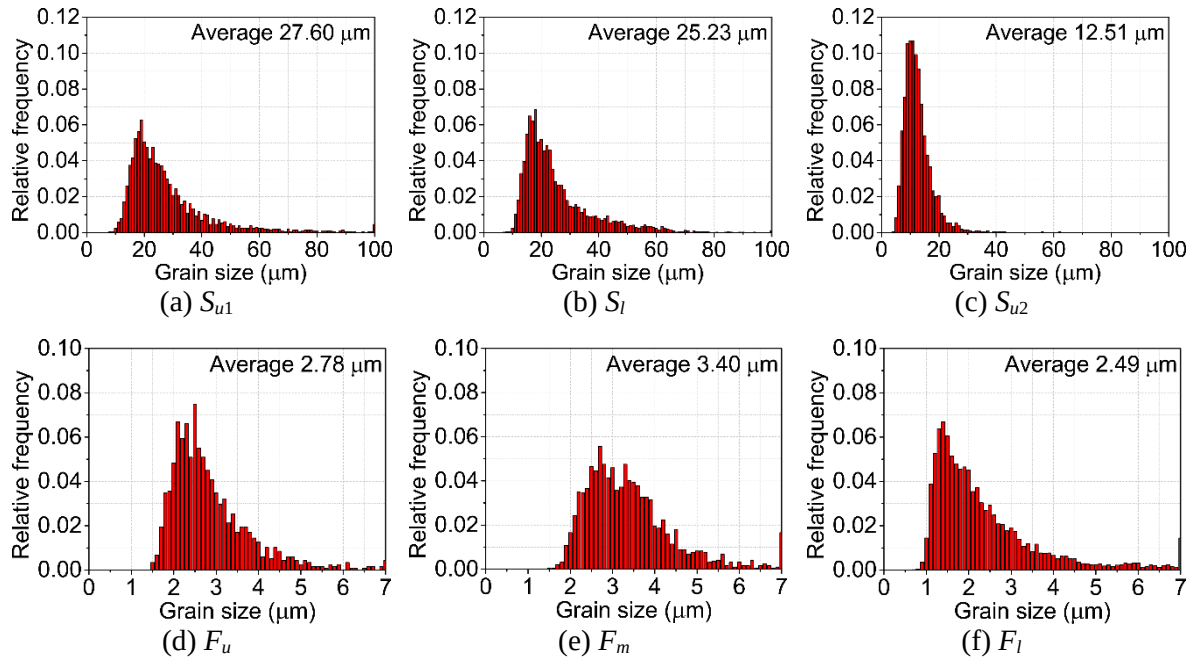
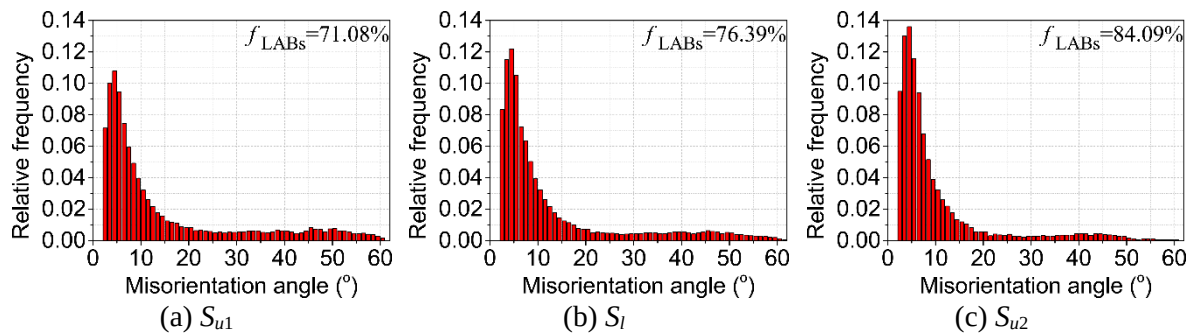


Fig. 6. Grain size distribution of the flow plane (a) shearing zone S_{u1} , (b) shearing zone S_l , (c) shearing zone S_{u2} , (d) outer part F_u of the formed curved bar, (e) middle part F_m of the formed curved bar, (f) inner part F_l of the formed curved bar.

Fig. 7 shows misorientation distribution of the grains shown in **Figs. 5b-g**. The f_{LABs} of the locations S_{u1} , S_l , S_{u2} in the shearing zone and F_u , F_m , F_l in the extrudate are 71.08%, 76.39%, 84.09%, 59.60%, 62.31%, and 49.76%, respectively. In **Figs. 7a** and **b**, a large number of

LABs which peak at $2\sim5^\circ$ appear at the start of shearing deformation. As shearing proceeds, the fraction of LABs increases dramatically due to evolvement of subgrain boundaries, especially for LABs within $2\sim5^\circ$, as shown in **Fig. 7c**. **Figs. 7d-f** show corresponding misorientation distribution of outer (F_u), middle (F_m) and inner (F_l) parts of the curved bar. Misorientation distribution peaks at $5\sim10^\circ$ at locations F_u , F_m , and F_l . It reveals a decrease in the frequency of $2\sim5^\circ$ boundaries accompanied by an increase in the frequency of HABs, as can be seen in **Figs. 7d** and **f**, another peak is found which is $44\sim48^\circ$ for F_u , and $51\sim55^\circ$ for F_l , respectively. The reduction of the fraction of LABs is due to further shear deformation, for F_u , along α -slip lines (flow lines) inside the central fan EPQ and along line EQ, as shown in **Fig. 1**. It is also noted that the f_{LABs} of F_l is lower than that of F_u and F_m , since the effective strain at F_l is larger than that at F_u and F_m , i.e. a greater strain leads to more LABs being transformed to HABs. The results reveal that refinement of grains in the curved AA1050 bar during DVSE at room temperature is due mainly to continuous dynamic recrystallization (cDRX), which involves formation of arrays of LABs and a gradual increase in the boundary misorientations during severe straining, finally leading to new grain development. Actually, both temperature and strain can affect recrystallization, and when strain increases recrystallization temperature is reduced. For high pure metals undergoing SPD, recrystallization temperature can be reduced below room temperature, and recrystallized structure may be observed after deformation [40].



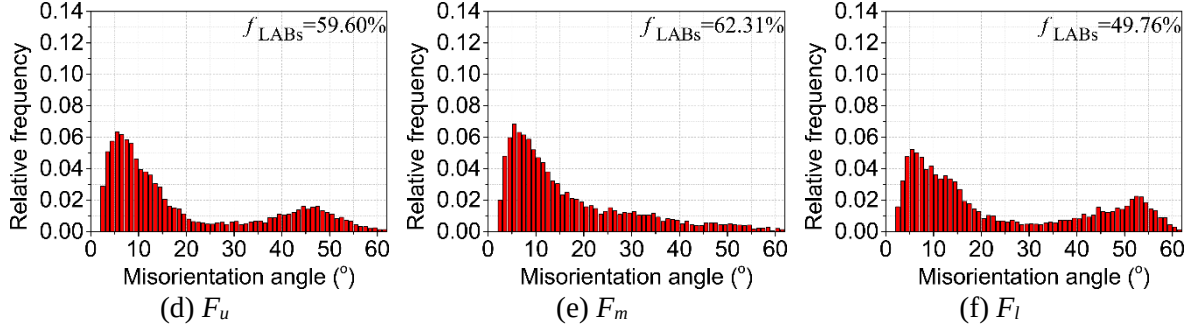
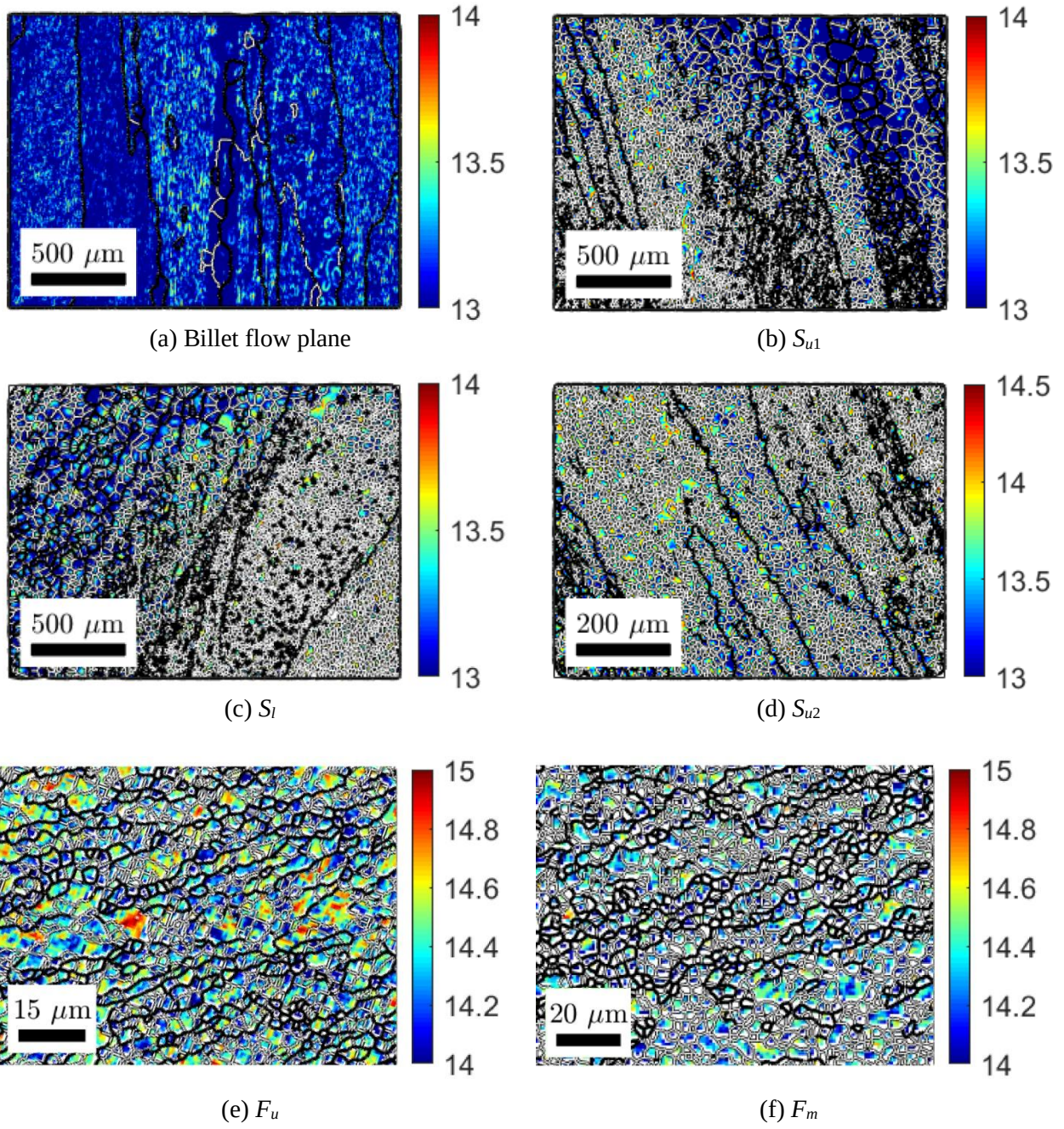


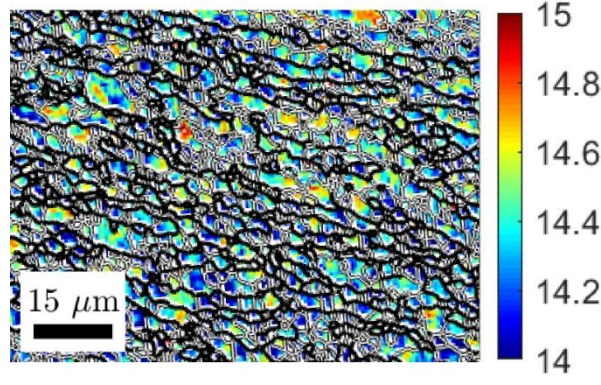
Fig. 7. Misorientation distribution of the flow plane (a) shearing zone S_{u1} , (b) shearing zone S_l , (c) shearing zone S_{u2} , (d) outer part F_u of the formed curved bar, (e) middle part F_m of the formed curved bar, (f) inner part F_l of the formed curved bar.

Cold plastic deformation in metals takes place in most conditions mainly by dislocation movement. The evolution of the dislocation density ρ_i inside the crystallites is assumed to take the form of KM equation $d\rho_i = (h-r\rho_i) d\varepsilon$, where h and r denote the strain hardening and dynamic recovery parameters [41]. When crystallites are deformed by a strain increment $d\varepsilon$, a dislocation density $d\rho_{i+} = h d\varepsilon$ is generated by straining, *i.e.* the athermal storage of dislocations due to moving dislocations blocked by obstacles, while $d\rho_{i-} = r\rho_i d\varepsilon$ is eliminated from ρ_i due to dynamic recovery either by ‘build’ into new LABs, or absorption into the pre-existent boundaries [42]. When a polycrystal is subjected to a deformation, strain heterogeneities occur in the grains, which are mostly near the grain boundaries. This happens especially for the SPD process whose main characteristic is the grain refinement. The refinement process of the grains involves generating new grain boundaries with increasing misorientations. These boundaries need to accommodate the deformation between neighbouring grains, and the related dislocations are termed geometrically necessary dislocations (GNDs) due to necessity for the geometrical accommodation [43]. GNDs comprise one part of the total dislocations, in which another part is ‘statistical’ dislocations. ‘Statistical’ dislocations do not lead to significant misorientations [44]. However, GNDs can produce misorientations greater than 15° , thus building up misorientations between the neighbouring grains.

The GNDs at different locations in the flow plane XY were calculated from EBSD measurements according to the procedure presented by Tóth et al. [43]: Firstly, to identify grain boundaries the grain tolerance angle was set as greater than 2° to examine the misorientation angle of each pixel with its neighbours, and the lattice curvature can be calculated based on the orientations of adjacent pixels. Then, from the lattice curvature, five components of the Nye dislocation density tensor can be obtained, and the GND scalar density can be defined as the entry-wise norm of the Nye dislocation density tensor divided by the Burgers vector length. The calculated GND density maps of the billet and selected locations S_{u1} , S_l , S_{u2} , F_u , F_m , F_l are shown in **Fig. 8**. Before DVSE, the billet has a very low GND density, as shown in **Fig. 8a**. As DVSE begins, the GND density is increasing, which is attributed to the increase of the deformation heterogeneity caused by the increase of the strain. As shown in **Figs. 8b-d**, SPD induces build-up of dislocation density that initiates subgrain walls having LABs at first, these are essentially dense dislocation cell walls whose misorientations are appreciably larger than those of common cell walls. As DVSE proceeds, dislocation density increases further which then is partly absorbed by the subgrain walls, leading to a transformation of LABs to HABs and thus a decrease of LABs and increase of HABs, as shown in **Figs. 8e-g**. The GND density increases up to an effective strain of ~ 2.75 (F_u , **Fig. 8e**), although there is an accompanied decrease of LABs when effective strain increases from ~ 1.0 (S_{u2} , **Fig. 8d**) to ~ 2.0 (F_m , **Fig. 8f**) and ~ 2.75 . This indicates that dynamic recrystallization starts when strain reaches ~ 1.0 , strain hardening is dominant over dynamic recovery and recrystallization until strain increases to ~ 2.75 . A slight decrease of GNDs arises at an effective strain of ~ 3.0 (F_l , **Fig. 8g**), where a further decrease in LABs, caused by the dynamic recrystallization, is also found. Actually, at very large strains, such as effective strains greater than 4~5 for fcc metal [43], dislocation density caused by SPD starts to decrease due to low dislocation accumulation capability of the submicron-sized and nano-

sized grains. More GNDs are absorbed by LABs and then transformed into HABs as a result of increased dynamic recovery and recrystallization, leading to a decrease of GNDs.





(g) F_l

Fig. 8. GND density maps of the billet and selected locations S_{u1} , S_l , S_{u2} and F_u , F_m , F_l . The colour scales are $\log_{10}$ scale, individually adjusted for each map to highlight dislocation structures.

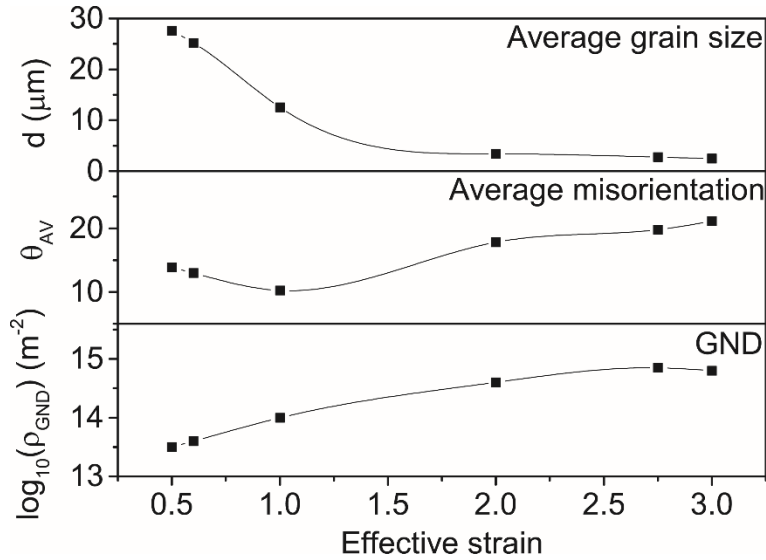


Fig. 9. Effect of DVSE effective strain on (i) average subgrain and grain size d , (ii) average grain boundary misorientation θ_{AV} , and (iii) GND density.

The evolution of microstructural parameters in the flow plane XY during DVSE are summarized in **Fig. 9**. This shows the dependence of average (sub)grain size d , average grain boundary misorientation θ_{AV} , and GND density ρ , on effective strain. It can be seen that the grain size drops rapidly at the start of straining (0~1.0). At locations S_{u1} , S_{u2} , S_l in the shearing zone, equiaxed grains with increasing LABs are generated in the interior of initial coarse grains (band-like structure with HABs), and an increased rotation of the band-like structure around the flow plane normal Z is found which is in accordance with the flow line.

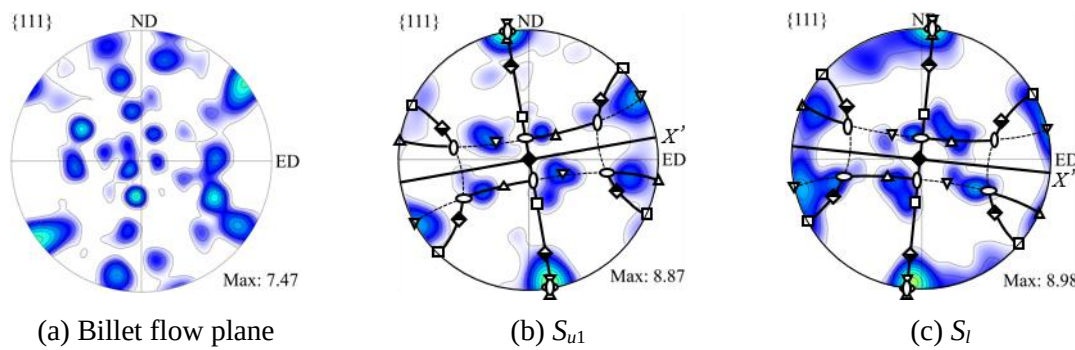
Angle θ_{AV} drops steadily at the start of straining (0~1.0) due to the formation of LABs, which is caused by the occurrence of dynamic recovery. GND density ρ increases progressively with strain due to the magnitude of the deformation heterogeneity increasing, which requires an increasing number of GNDs. At locations F_u , F_m , F_l in the extrudate, it can be seen that grain size approaches a constant value of $\sim 3 \mu\text{m}$ at high effective strain ~ 2.0 . The size of the band-like structure (elongated grains) with HABs is greatly reduced, compared with that of the shearing zone, the elongated grains are aligned approximately along the shear direction of the ideal intersection planes. Equiaxed grains with increasing HABs are generated within elongated grains. Angle θ_{AV} rises steadily with straining (1.0~3.0), due to the transformation of LABs to HABs. The increase of HABs is due to the occurrence of continuous dynamic recrystallization (CDRX). GND density ρ increases progressively with strain up to an effective strain of ~ 2.75 , followed by a steady decrease at larger strains. In conclusion, in the DVSE process, initial coarse grains are firstly subdivided into equiaxed subgrains with LABs, then fine grains with HABs are generated and elongated along shear direction of the ideal intersection planes. The grain refinement mechanism is CDRX.

3.2. Micro-texture

Fig. 10 shows the $\{111\}$ pole figure of the original billet and different locations in the flow plane of the shearing zone and curved bar during the DVSE process, where ED and ND indicate the extrusion direction and normal direction. For comparison, **Figs. 10h-i** show the ideal position of some typical texture components in $\{111\}$ pole figure for fcc materials after negative simple shear (shearing strain $\gamma=-2$) and one pass of ECAP deformation with 90° channel angle, respectively [45]. The negative simple shear means the shear direction is along the $-X'$ direction, as shown in **Fig. 10h**. For ECAP with 90° channel angle, the apparent shear plane $X'Z'$ (normal Y') in **Fig. 10i** has an anticlockwise angle $\theta=45^\circ$ to the extrusion direction $+X$ (ED), and so the ideal position of typical texture components needs to be rotated by an

anticlockwise angle $\theta=45^\circ$. Typical ideal shear texture components and their corresponding Miller indices are given in **Table 1**.

The scattered distribution and the low maximum density shown in **Fig. 10a** illustrate that the original billet has very weak texture components. A-type shear-texture components A_2^* ($\{11\bar{1}\}\langle 112\rangle$) and A_1^* ($\{\bar{1}\bar{1}1\}\langle 112\rangle$), are formed at locations S_{u1} and S_l in the shearing zone. Besides A-type shear-texture components, C-type shear-texture component C ($\{001\}\langle 110\rangle$) is also formed in S_l , although it is weaker than that of A-type. The apparent shear plane $X'Z'$ (normal Y') in **Figs. 10b-c**, has angle $\theta_{u1}\sim 10^\circ$ and $\theta_l\sim 5^\circ$ to the extrusion direction $+X$ (ED), respectively. Angle θ_l is lower than θ_{u1} due to greater effective extrusion ratio and severe shearing in zone S_l . As shear strain increases, the intensity of the C-type shear-texture component C increases significantly, as can be seen in **Figs. 10d-g**. The apparent shear plane $X'Z'$ (normal Y') in **Fig. 10d**, has angle $\theta_{u2}\sim 15^\circ$ to the extrusion direction $+X$ (ED), due to its rotation around Z axis (TD) as the shearing process proceeds. At locations F_u , F_m , and F_l in the extrudate, C-type shear-texture component C is the main texture, and the shear-texture intensity increases from F_m to F_u , and F_l , due to increase of the effective strain. The angle between the apparent shear plane $X'Z'$ (normal Y') and the extrusion direction $+X$ (ED) is the minimum $\theta_{fl}\sim 18^\circ$ for F_l , followed by $\theta_{fu}\sim 22^\circ$ for F_u and $\theta_{fm}\sim 25^\circ$ for F_m , which is in accordance with the extent of the effective extrusion ratio and effective strain.



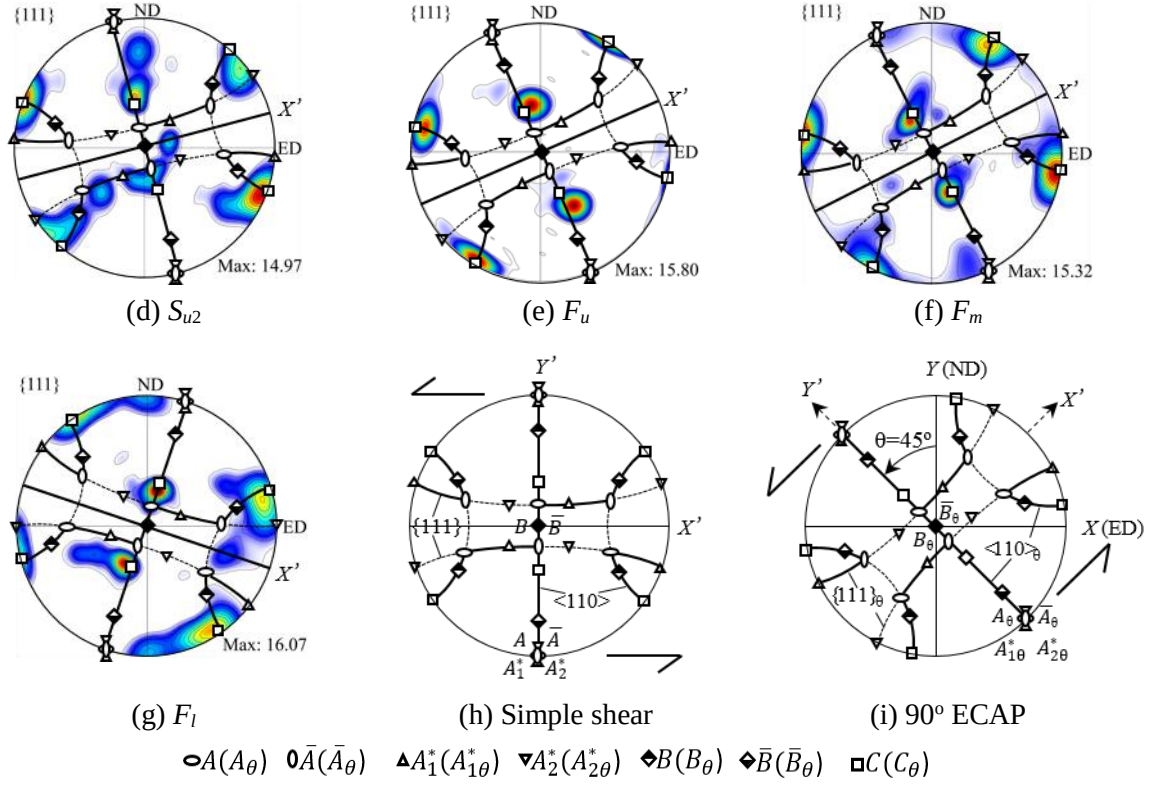


Fig. 10. The $\{111\}$ pole figures of the original billet and different locations in the flow plane of the shearing zone and curved bar.

Table 1 Typical ideal shear texture components with the corresponding Miller indices

Components	A	$\bar{A}$	A_1^*	A_2^*	B	$\bar{B}$	C	A fibre	B fibre
Miller indices	$\{1\bar{1}\bar{1}\}$ $\langle 110 \rangle$	$\{\bar{1}11\}$ $\langle \bar{1}\bar{1}0 \rangle$	$\{\bar{1}\bar{1}1\}$ $\langle \bar{1}12 \rangle$	$\{11\bar{1}\}$ $\langle 112 \rangle$	$\{\bar{1}12\}$ $\langle \bar{1}10 \rangle$	$\{1\bar{1}\bar{2}\}$ $\langle \bar{1}\bar{1}0 \rangle$	$\{001\}$ $\langle 110 \rangle$	$\{111\}$ $\langle uvw \rangle$	$\{hkl\}$ $\langle 110 \rangle$

The orientation distribution function (ODF) sections of $\varphi_2=45^\circ$ for locations S_{u2} , F_u , F_m and F_l , are shown in **Fig. 11**. The ODF sections for S_{u1} and S_l are not given due to the low intensity of texture. Locations of the main ideal orientations of textures in the ODF section of $\varphi_2=45^\circ$ for fcc materials are also given in **Fig. 11**. Shown in **Fig. 11a** is the case of negative simple shear (shearing strain $\gamma=-2$) and in **Fig. 11f** the case for one pass of ideal ECAP deformation with 90° channel angle [45]. As indicated by the dashed lines, the main components in the DVSE texture (**Figs. 11b-e**) are shifted along the φ_1 axis by different angles from those in simple shear texture (**Fig. 11a**), as occurs for ECAP texture which can

be obtained by a shift of 45° (half channel angle) from the simple shear texture. This relation holds for the entire texture. The shifted angles are approximately the same as the angles between the apparent shear plane $X'Z'$ (normal Y') and extrusion direction $+X$ (ED) in **Figs. 10d-g**, which are $\theta_{u2} \sim 15^\circ$ for S_{u2} , $\theta_{fl} \sim 18^\circ$ for F_l , followed by $\theta_{fu} \sim 22^\circ$ for F_u and $\theta_{fm} \sim 25^\circ$ for F_m .

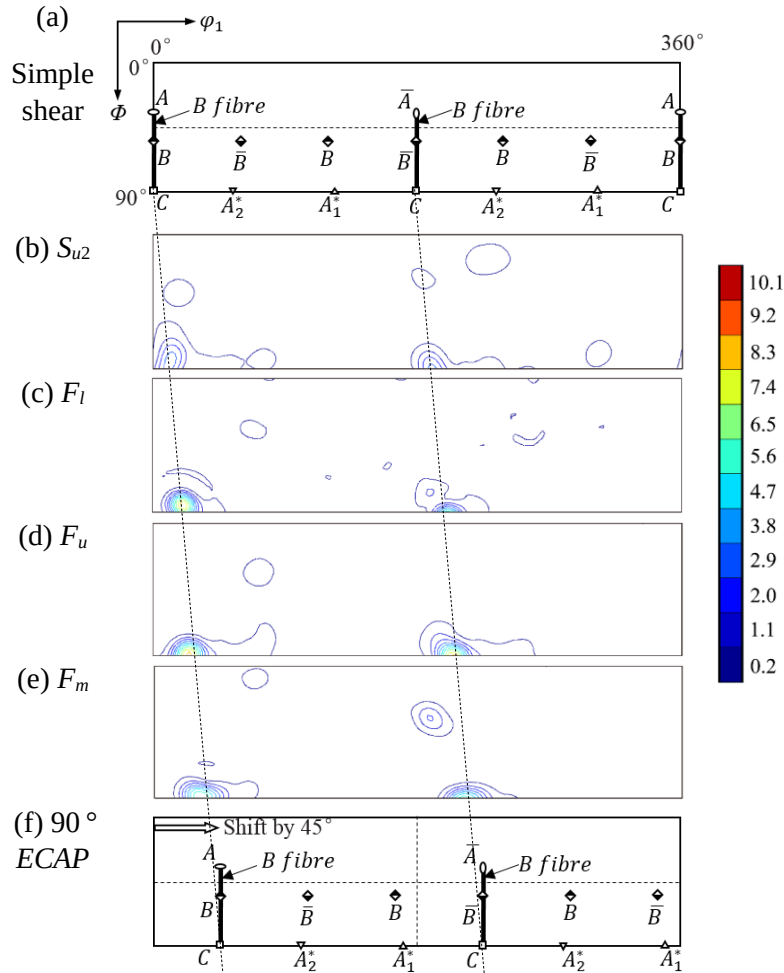


Fig. 11. The orientation distribution function (ODF) sections of $\varphi_2=45^\circ$ for locations S_{u2} , F_u , F_m , F_l .

3.3. Micro-hardness and tensile properties

Micro-hardness (H_V) of the billet and selected locations S_{u1} , S_l , S_{u2} and F_u , F_m , F_l is shown in **Fig. 12**. During DVSE, the hardness gradually increases by 134.8% from 23 HV to 54 HV, which is consistent with the values of local effective strain. Hardness increases as grain size decreases, and the relationship follows approximately the Hall-Petch equation:

$$H_V = H_0 + K_H d^{-1/2} \quad (1)$$

where H_0 and K_H are material constants, calculated to be 23.68 and 50.38 respectively.

Fig. 13 shows the room temperature engineering stress-strain curves of tensile samples cut from the billet and DVSE-processed (DVSEd) curved bar. The 0.2% proof stress (yield strength YS), ultimate tensile strength (UTS) and elongation to fracture (EL) were calculated from the above stress-strain curves. The YS increases from 29.8 to 135.3 MPa and the UTS from 68.6 to 148.7 MPa, although the EL decreases by 60.0% from 0.55 to 0.22. The YS and UTS of the DVSEd bar have been improved by 354.0% and 116.8%, respectively. Such a significant increase in YS, UTS and decrease in plasticity are a consequence of grain refinement caused by severe plastic deformation and an increase in dislocation density. An increased volume fraction of grain boundaries as a result of grain refinement leads to an increase in the number of barriers to the movement of dislocations, therefore an increase in YS and UTS while a decrease in plasticity are achieved.

As pointed out by Azushima et al. [46], the decrease in ductility is a general feature of strain-hardened metallic materials. Thus, it is not surprising that the SPD-processed materials show limited tensile ductility. An increased volume fraction of grain boundaries in SPD-processed materials leads to the reduced ability of dislocation movement and increased difficulty in further accumulation of dislocations. As a result, the SPD-processed material has a limited strain hardening capability, which leads to immediately plastic instability and the failure of the sample. As shown in **Fig. 13**, the billet exhibits a stress-strain curve that indicates normal strain hardening, while the DVSEd curved profile hardly exhibits strain hardening.

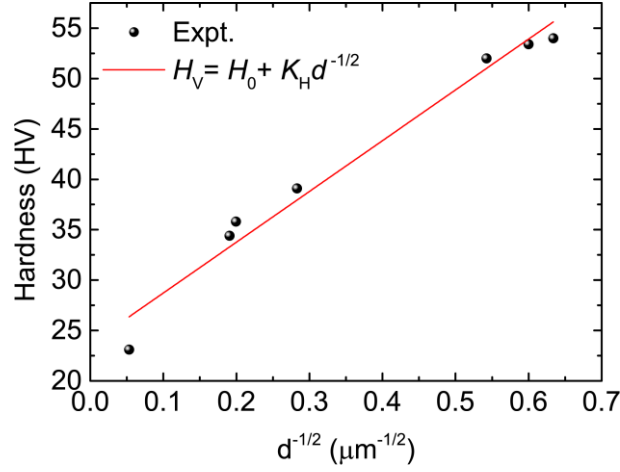


Fig. 12. Micro-hardness of the billet and selected locations S_{u1} , S_l , S_{u2} and F_u , F_m , F_l .

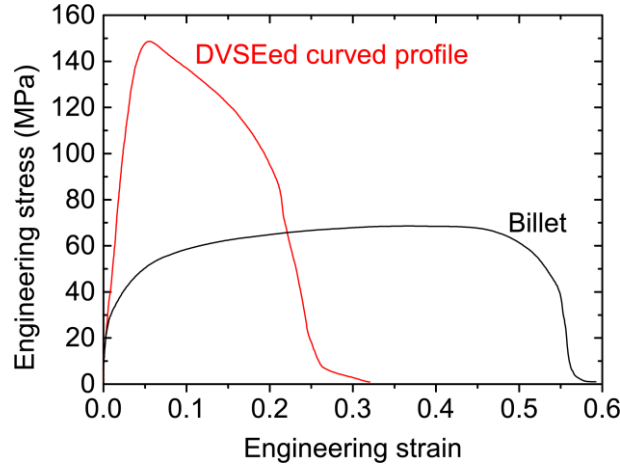


Fig. 13. Tensile properties of samples from the billet and DVSEed bar.

4. Conclusions

In the present work, a novel curved profile forming process, differential velocity sideways extrusion (DVSE), was used to form a curved bar of aluminium alloy AA1050 at room temperature. The evolution of grain structure and micro-texture of AA1050 during DVSE were studied, and the grain refinement mechanism was revealed. Mechanical properties of the formed curved bar using DVSE were also examined. The following major conclusions are drawn:

- (1) DVSE process not only can be used to form a curved bar of AA1050 at room temperature in one operation, but also leads to a remarkable grain refinement, from $\sim 357 \mu m$ (average

grain size) of the original billet to $\sim 3 \mu\text{m}$ (average grain size) of the curved bar. That is, grains are refined to less than 1% of their original size.

(2) The significant refinement of AA1050 grains in the DVSE process is mainly due to the continuous dynamic recrystallization (cDRX), which is attributed to the SPD occurring at intersecting regions. DVSE with an extrusion ratio of 1.6 incurred an effective strain of ~ 2.0 in the extrudate, which is greater than the value of ~ 1.0 obtained using one pass ECAP with 90° channel angle. Due to greater effective strain, slightly greater grain refinement and high angle boundary fraction were found in the inside bend of the curved bar than at the outside.

(3) During DVSE, coarse band-like structures containing subgrains with LABs were found at the start of the shearing zone, which then transform into fine shear band-like structures having HABs. The band-like structures are inclined approximately along the intersection of shear planes. Within these structures are equiaxed (sub)grains having a mixture of LABs and HABs. A-type shear-texture components are generated in the shearing zone, and they transform into the C-type component after DVSE. The texture can be derived from those for negative simple shear texture by shifting along the ϕ_1 direction with an angle approximately equal to the angle between the intersection plane and extrusion direction.

(4) Compared with the billet, grain refinement and high dislocation density in the curved bar are associated with an increase in hardness, yield strength and ultimate tensile strength by 134.8%, 354.0% and 116.8% respectively, while the elongation to fracture is decreased by 60.0%.

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