

Micro and Macro Analysis of Anisotropy of an AA3104 Aluminum Alloy

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Abstract. Anisotropy plays an important role when forming aluminum alloys. Measurement of anisotropy using traditional extensometers can be challenging when working with AA3104 alloys due to the presence of Piobert-Lüder's banding and the Portevin-Le Chatelier effect. Piobert-Lüder's bands creates large variations in r-value necessitating alternative analysis techniques. In the present work r-values obtained from digital image correlation are compared to viscoplastic self-consistent (VPSC) model predictions. The VPSC predictions are computed based on experimental texture measurements. Good correlation was obtained between the digital image correlation measured and crystal plasticity predicted r-values. The r-value results are assessed based on contribution of different crystallographic texture components.

1. Introduction

Forming components from rolled aluminum sheet is advantageous due to the favorable density, strength, formability, and recyclability of the material. Many products are currently fabricated from rolled aluminum sheet, such as, but not limited to: beverage cans, food containers, automotive skin panels, and electronic chassis components. A common issue when forming these components is the inherent anisotropy present in the material due to texture development during rolling. This anisotropy should be accounted for during design to ensure correct material flow during manufacturing resulting in desired as-formed geometry. Finite element simulations are commonly used during design and require advanced material models to account for anisotropy, such as those developed by Barlat *et al.* [1,2], Banabic *et al.* [3,4], Hill [5], and Vegter [6]. Common inputs to these models are flow stress ratios and Lankford coefficients (r-values) derived from standard tensile data from multiple orientations to the sheet rolling direction. Some of these models require minimal tensile testing (tensile tests in 3 directions), while others require more extensive testing (e.g. tensile tests every 15°). Some of the models also require data from tests other than tensile conditions, e.g. biaxial, plane strain, and shear.

Unfortunately, some dilute aluminum alloys containing magnesium and/or manganese are affected by the Portevin-Le Chatelier (PLC) [7] effect and Piobert-Lüder's (Lüder's) banding [8,9] under certain loading and temperature conditions. These effects create serrated flow and non-uniform strain fields leading to, among other things, r-value measurement difficulties, especially with traditional extensometers. Alternative methods can be used to overcome these measurement difficulties, such as digital image correlation (DIC) and predictions from viscoplastic self-consistent (VPSC) models. As such, the purpose of the present work is to investigate anisotropy of a rolled aluminum sheet affected



by non-uniform strain distribution using DIC and VPSC. DIC data is used to investigate r-value variation resulting from the non-uniform strain distribution and is compared to r-values predicted from VPSC calculations. Note that VPSC is being used in the present work to predict the bulk response of the material and not to predict non-uniform strain distribution. Predicted texture evolution is also used to help explain the DIC observed strain distributions. A beverage can aluminum alloy (3104) cold rolled to 0.258mm thick is chosen for this work. The material is tested in the extra hard condition (H19) and is the typical starting condition for beverage can manufacturing. This alloy nominally contains 0.8 – 1.4% manganese and 0.8 – 1.3% magnesium producing good conditions for Lüder's band development.

2. Viscoplastic Self-Consistent Model

The VPSC model describes the plastic response of a polycrystal under imposed boundary conditions [10]. It makes use of the Eshelby [11] solution, originally used for treating the elastic problem, by linearizing the non-linear plastic response of the polycrystal and its grains in incremental form. For this purpose, each grain of the polycrystal is represented as an ellipsoidal inclusion embedded in a homogeneous equivalent medium (HEM) representing the polycrystal itself. The constitutive relationship for the individual grains is written as

$$\dot{\epsilon}_{ij} = M_{ijkl} \dot{\sigma}_{kl},$$

where $\dot{\epsilon}_{ij}$ is the grain strain rate, M_{ijkl} is the grain compliance, and $\dot{\sigma}_{kl}$ is the grain deviatoric stress. Similarly, the polycrystal constitutive relationship is expressed as

$$\bar{\dot{\epsilon}}_{ij} = \bar{M}_{ijkl} \bar{\dot{\sigma}}_{kl},$$

where $\bar{\dot{\epsilon}}_{ij}$, $\bar{M}_{ijkl}$, and $\bar{\dot{\sigma}}_{kl}$ now refer to the HEM. Utilizing the Eshelby solution, the grain level stresses and strain rates are related to the polycrystal values through the interaction equation, i.e. [10,12]

$$(\dot{\epsilon}_{ij} - \bar{\dot{\epsilon}}_{ij}) = -\tilde{M}_{ijkl}(\dot{\sigma}_{kl} - \bar{\dot{\sigma}}_{kl}).$$

Here, $\tilde{M}_{ijkl}$ is the interaction tensor [13,14] which itself is defined as

$$\tilde{M}_{ijkl} = n^{\text{eff}}(I_{ijmn} - S_{ijmn})^{-1} S_{mnpq} \bar{M}_{pqkl},$$

where n^{eff} is a parameter controlling the strength of interaction, I_{ijmn} is the identity tensor, and S_{mnpq} is the Eshelby tensor. To solve for the evolution of the macroscopic polycrystal (HEM) stress when a macroscopic strain rate is imposed on the HEM, the iterative self-consistent process is deployed. Further detail on the self-consistent algorithm may be found in the original work of Lebensohn and Tome [10].

In this work, we used the experimentally measured crystallographic texture of the material to assign the anisotropic behavior of the polycrystal. For this purpose, the measured texture was represented by 5000 discrete Euler angle triplets, i.e. the grains of the polycrystal. Thereafter, a tensile velocity gradient was imposed on the aggregate. The velocity gradient corresponded to a strain rate of 0.001s^{-1} . The aggregate was pulled in tension to a von Mises strain of 0.2 and subsequently the r-value was calculated. The angular variation of the r-value with respect to the rolling direction was determined by incrementally rotating the crystallographic texture around the normal direction of rolling and performing the r-value calculation again. Noting that throughout the entire analysis in this work, the critical resolved shear stress of the slip planes, τ_0 , was set to a value of 1 and the other parameters associated with the extended Voce hardening law on the slip planes were set to zero. Various

hardening parameters and models were analyzed and it was found that the calculated r -values were independent of the hardening behavior on the slip plane.

3. Experimental Methods

Tensile tests were performed using full-size specimens manufactured according to ASTM E517 [15] resulting in reduced section width and length of 12.5mm and 75mm respectively. Gauge width and length were chosen to be 8mm and 50mm respectively. Specimens were extracted every 15° to the rolling direction and 5 repeat tests were performed for all conditions. All testing was performed on an MTS electromechanical load frame with a 2.5kN load cell and standard wedge-action grips. A crosshead speed of 0.05mm/s was selected resulting in an initial quasi-static strain rate of 0.001s^{-1} . A 3D GOM Aramis 4M DIC system was used with 50mm lenses. Images, load, and crosshead displacement were recorded by the Aramis system at a rate of 5 Hz. Facet size, facet step, and strain computation size were 20, 10, and 3, respectively. r -values were determined using two methods from the calculated DIC data. The first method simulated traditional extensometers using the DIC software by placing virtual extensometers to measure changes in length and width. Clip-on extensometers were not available and would have been difficult to use considering the thin specimens. The second method involved calculated r -values for every DIC point and averaging the results over the entire reduced section of the tensile specimen. X-Ray diffraction (XRD) was used to measure textures for input to VPSC. Specimens were ground using 600-grit paper down to $\frac{1}{4}$ of sheet thickness and washed with nitric acid. A Panalytical Empyrean XRD was used with a polycapillary lens and point detector. Four incomplete pole figures were measured ([111], [200], [220], and [311]) and the orientation distribution function (ODF) was calculated using mTex [16].

4. Results

The challenges associated with r -value measurement of the material under investigation become evident when analyzing the engineering stress-strain curves, as shown in Figure 1 (a and b). Figure 1 (a) shows 5 repeat tests of the measured engineering stress-strain response in the 0° (rolling) direction with excellent repeatability between all tests. The high strength and low tensile elongation of the material are a result of the H19 condition. It is immediately clear that strain non-uniformity is present based on the jumps in stress at engineering strains of approximately 2% and 4%. Other tensile tests using different test machines and measurement equipment have also revealed these stress jumps. The repeatability of the stress jump is striking as demonstrated by the inset of Figure 1 (a) that shows a magnified region surrounding the first stress jump. It can be seen from this magnified region that the stress jumps nearly instantly by approximately 5 MPa (or nearly 2%). Figure 1 (b) shows single representative curves for all of the tested directions. Repeatability of all stress-strain curves was excellent and on-par those shown in Figure 1 (a). Tests performed in the 0° , 15° , 30° , and 45° directions display similar overall stress-strain responses; however, the stress jump appears less sudden for the 15° , 30° , and 45° directions as compared to the 0° direction. Overall stress increases for the 60° , 75° , and 90° directions compared to the other directions and the stress jump became less sudden and also shifted to slightly higher strains.

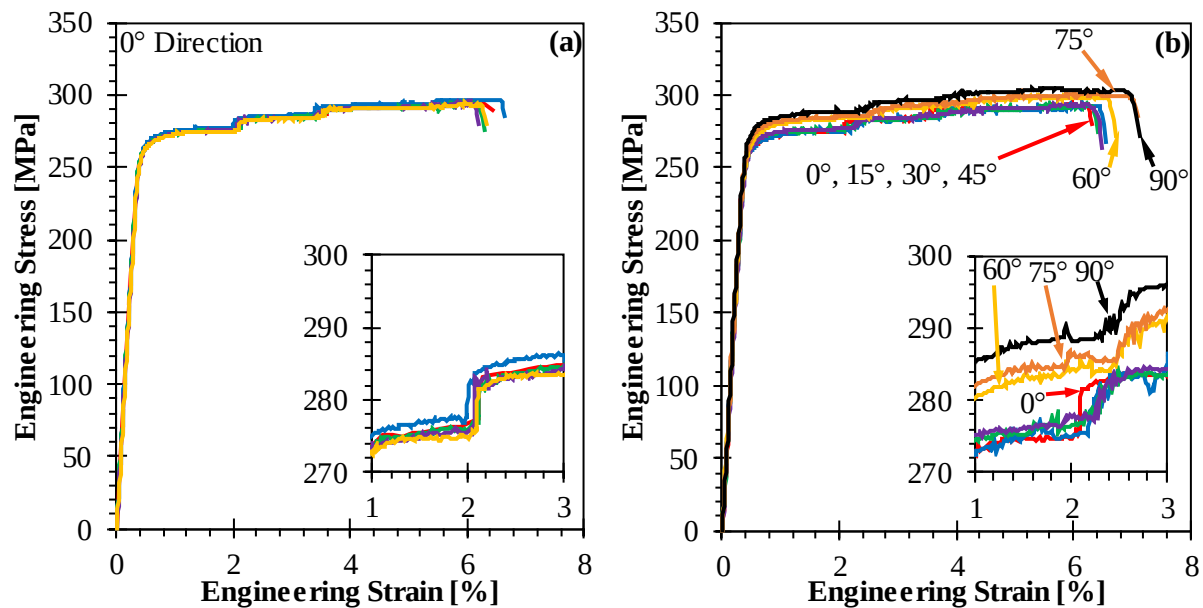


Figure 1. Measured engineering stress - strain curves. a) repeat tests in rolling (0°) direction, and b) summary of representative tests in all directions.

The change in the stress jump as the test direction changes from 0° to 90° is quite interesting and may be partially explained by Figure 2, which shows the DIC calculated von Mises strain contours for all tested directions at the same tensile elongation of 3%. The change in the contours seems to suggest that Lüder's band formation is quite regular when the test direction is aligned with the sheet rolling direction and becomes more stochastic and with multiple initiation sites as the test direction is rotated away from the sheet rolling direction. Indeed, videos of the strain contours reveal a single band forming and traversing the length of the specimen in the 0° direction whereas multiple bands form and interact at angles greater than 0° .

Clearly, strain non-uniformity is a concern as demonstrated in Figure 2 and has severe consequences when attempting to determine r -values. Figure 3 (a and b) shows r -values obtained from simulated extensometers and the area average method. It can be seen in Figure 3 (a) that simulated extensometers create erratic results and clearly one can not specify an r -value. Although a trend of increasing r -value with increasing test angle is observable, there is considerable overlap due to the sawtooth shape of the curves. The repeatability of the r -value measurements became increasingly poor as the direction increased from 0° to 90° . At 0° the peaks and valleys nearly lined up, but this was not the case in the other test directions. The lack of repeatability in r -values at higher testing angles is likely related to the increased random initiation of Lüder's banding at higher testing angles. Figure 3 (b) shows that the values obtained from the area average method are smooth and the repeatability of the area average method was excellent with the curves from 5 tests lining up on top of each other. Using the area average method more clearly shows r -values that are nearly constant with increasing strain and a trend of increasing r -value with increasing test angle. The area average method is, in essence, a low-pass filter on the data.

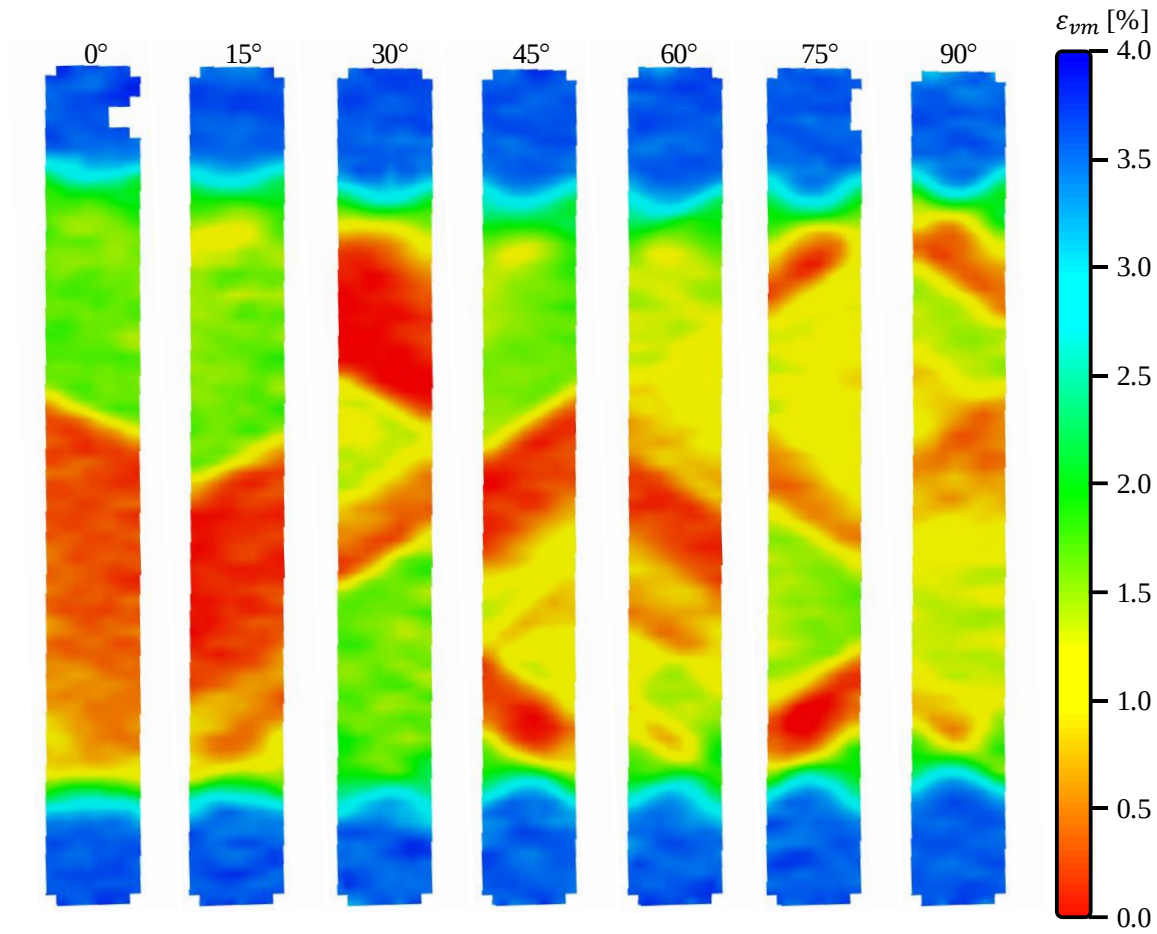


Figure 2. Measured von Mises strain contours from DIC for each tested direction at 3% elongation.

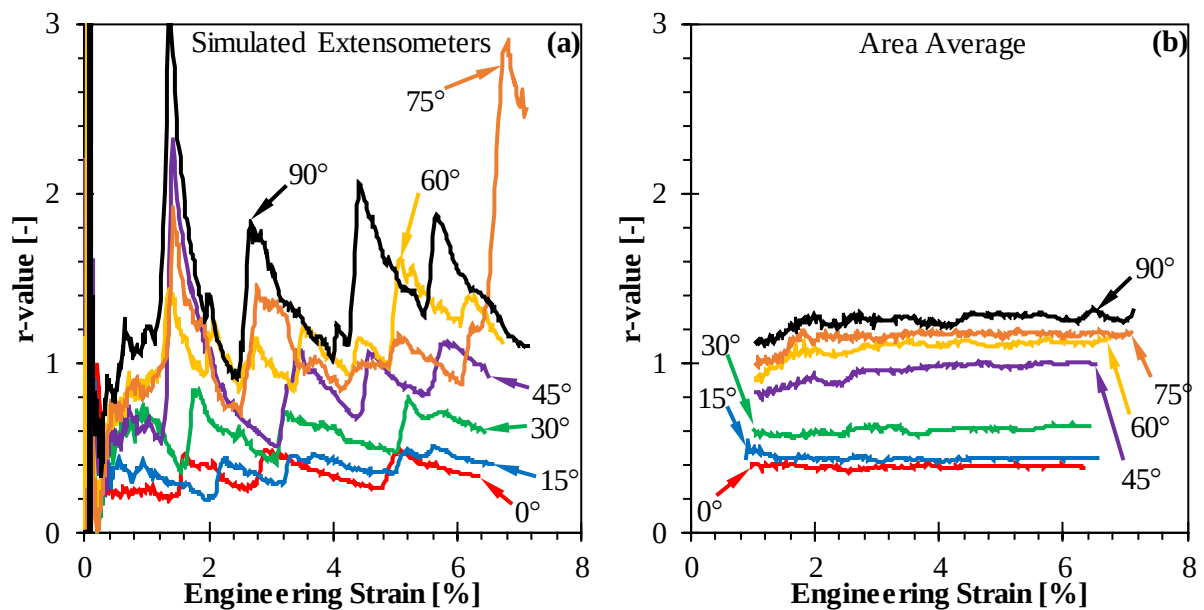


Figure 3. Representative r-values as a function of engineering strain for each tested direction. a) values from simulated extensometers, and b) values from area average.

The alternative to measuring r -values from tensile tests is to predict the values from VPSC calculations. Figure 4 demonstrates this prediction. The average r -value (calculated between 1% – 5% engineering strain) from 5 repeat tests using simulated extensometers and the area average method are shown as circle and diamond symbols, respectively. These averaged results are in reasonable agreement near 0° from the rolling direction; the agreement and the error of the simulated extensometers becomes worse at higher angles. VPSC predictions are shown as solid lines and dashed lines for the edge and center of the coil, respectively, with each of the three curves representing the head, middle, and tail of the coil. The VPSC predicted r -values show good agreement with the measured r -values for the edge of the coil and is a positive finding as the tensile specimens and XRD edge specimen were extracted from similar locations on the coil. The predicted r -values from the center of the coil show a different trend due to the different thermo-mechanical processing that occurs across a coil.

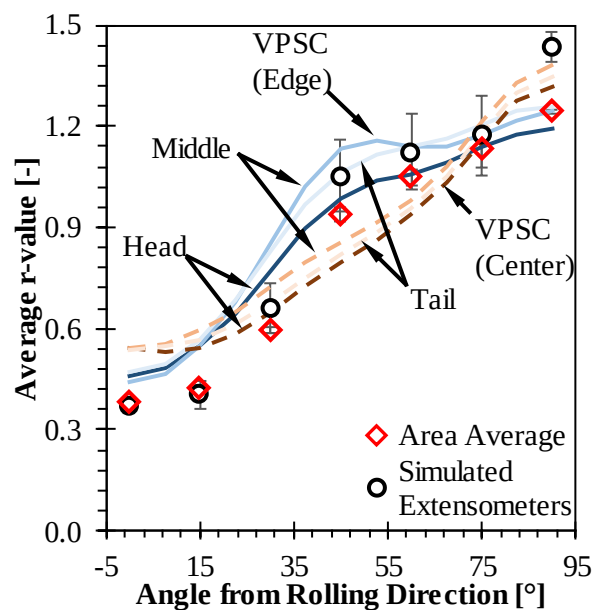


Figure 4. Comparison of DIC measured and VPSC predicted r -values as a function of angle from the rolling direction.

Figure 5 (a, b, and c) shows predicted pole figures of select orientations at several strain levels. The figure demonstrates the change in texture when the initial rolling texture is strained in tension at different directions to the rolling direction. In Figure 5 (a) the texture does not seem to change with increased strain and can likely be attributed to the alignment of the tensile loading direction with the strong rolled-in texture. Texture evolution is more noticeable when the loading direction is aligned at 45° and 90° to the rolling direction as shown in Figure 5 (b and c), where the change appears more apparent as the angle from the rolling direction is increased.

Figure 6 (a, b, and c) shows the calculated texture components as a function of strain for tensile loading directions of 0°, 45°, and 90°. The loading directions in Figure 6 are the same as those in Figure 5. Figure 6 (a) reinforces the results from Figure 5 (a) and shows the dominant rolled-in texture (S-component) does not change appreciably with increasing strain. Again, this may be attributable to the S-component being aligned with the tensile rolling direction. The percentage of Copper-component increases the most of all texture components. When pulled in the 45° direction the percentage of S-, Brass-, and Cube-component do not increase much, but the Copper-, Cube_{RD}-, and Goss-component do increase. Also note that the Cube-component percentage is nearly zero. When pulled in the 90° direction the percentages of Copper- and Goss-component are nearly zero with very little change with

increasing strain and the other components have nearly the same initial and increases in percentages with increasing strain.

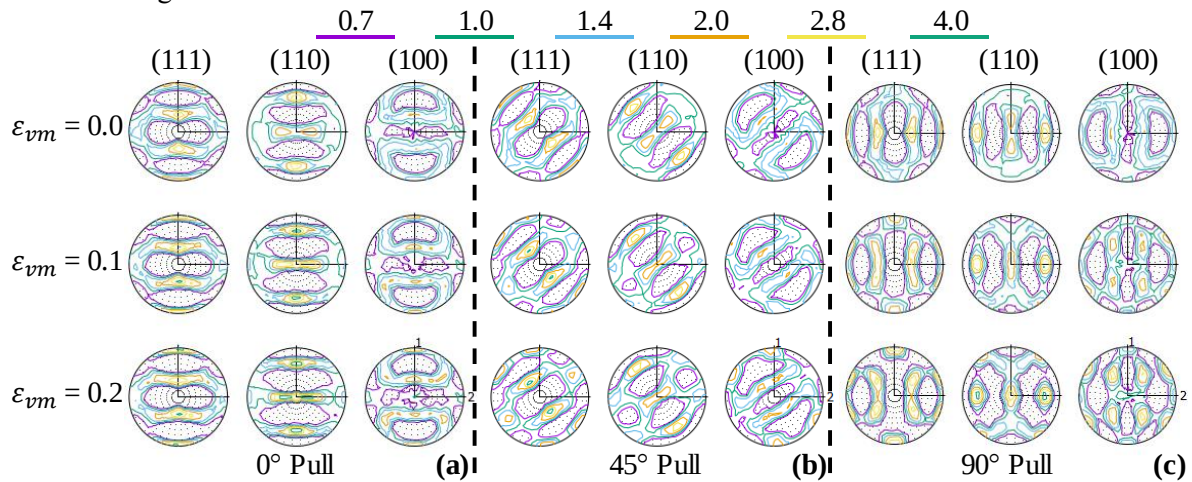


Figure 5. Pole figures of select orientations as a function of von Mises strain for three tensile pull directions. a) 0° direction, b) 45° direction, and c) 90° direction.

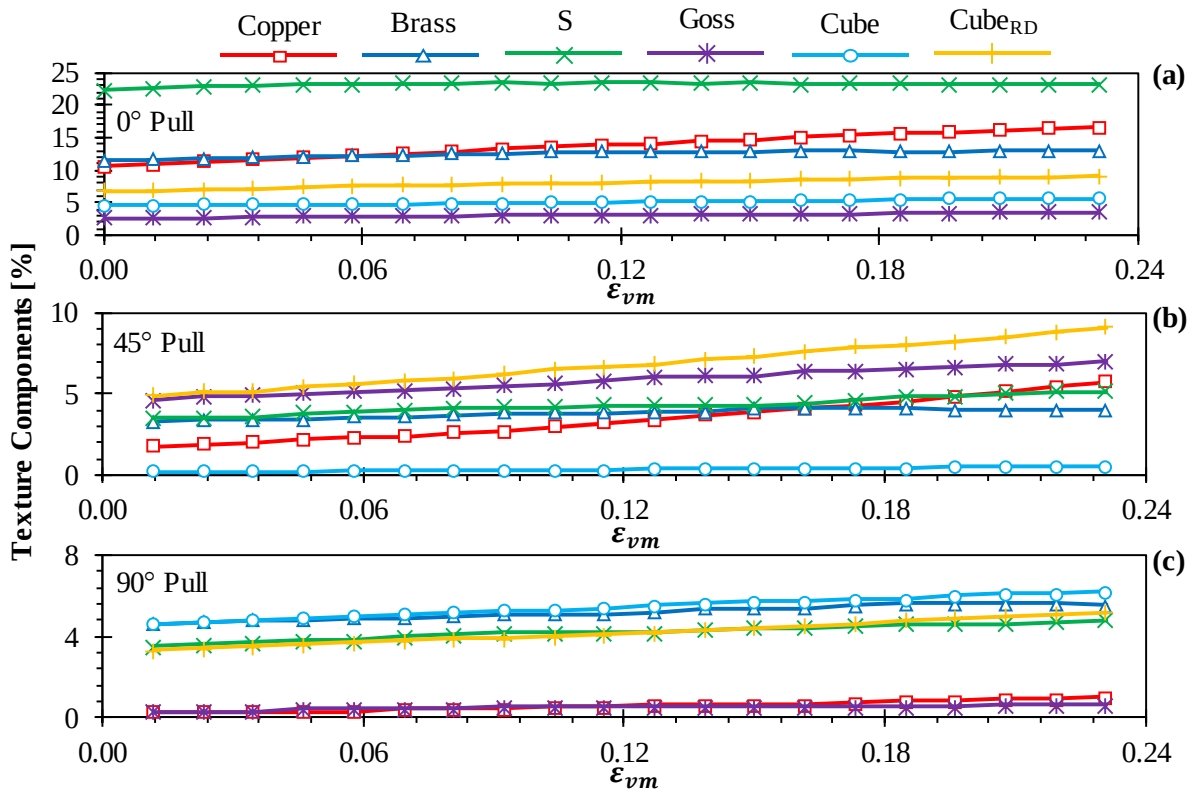


Figure 6. Percent evolution of select texture components as a function of von Mises strain for three tensile pull directions. a) 0° direction, b) 45° direction, and c) 90° direction.

5. Discussion

The preceding results show that it can be difficult (or impossible) to measure r-values of materials susceptible to the PLC effect and Lüder's banding using conventional extensometers, but r-values can be determined using DIC and averaging results over the entire reduced section of the specimen.

Unfortunately, DIC can be more time consuming due to the required setup and data analysis rigor and the data archiving requirements. The alternative demonstrated above utilizes VPSC predictions using experimentally measured textures and this method closely matches the DIC measured results. VPSC and texture measurements require significantly less analysis and specimen preparation time permitting sampling at multiple coil locations leading to a better understanding of anisotropy variation throughout the coil. VPSC predictions also allow for determination of r -values at many different angles to the rolling direction leading to additional data for yield surface development.

Throughout this work the origin and evolution with respect to rolling direction of the observed stress jumps was puzzling. Using VPSC calculations to predict texture evolution and the active texture components shed some light on the phenomenon. The regularity of the banding and stress jumps when specimens were tested in the 0° direction may be attributable to the strong rolled-in texture and dominance of S-component. Essentially, pulling in the rolling direction does not sufficiently activate other texture components resulting in repeated Lüder's band formation and stress jumps once the entire gauge region is consumed by the band. It may be that other texture components become more active as the test direction deviates from the rolling direction leading to texture evolution and random Lüder's band formation, or a transition from Lüder's banding to PLC effect. Naturally, this is somewhat speculative and requires further study and analysis.

Additional work to determine the cause and origin of the stress jumps is required. The effect of specimen geometry, stress state (e.g. biaxial, plane strain, shear), temperature, and test speed should be investigated. Strain rate jump tests may be useful to determine the strain rate sensitivity of the investigated material and its relation to Lüder's band formation. In depth optical, scanning, and tunneling electron microscopy may reveal microstructural origins and the relation of dislocation movement and pinning to the phenomenon observed in the present work.

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