Polycrystal Simulations
Investigating the Effect of Additional Slip System Availability in a 6063 Aluminum Alloy at Elevated Temperature

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1 Introduction

A better understanding of the relationship between processing and product properties can help to improve the performance of aluminum alloys. The macroscale properties that influence performance, such as formability, yield strength, and anisotropy, are largely dictated by microstructural features including grain shape, grain size, and orientation and misorientation distributions, which are formed during the metal forming process (see Olaf and Hirsch [1] for a review).

Grains distort and rotate to accommodate deformation. The level of stretching and rotation during deformation is related to the amount of slip along crystallographic slip systems within the grain, which, in turn, is related to dislocation motion. The rotational component of the deformation tends to cause grains to rotate toward preferred crystallographic orientations, depending on the nature of the deformation, resulting in a pattern of crystallographic orientations referred to as the texture.

The joint requirements of compatibility and stress equilibrium at grain boundaries lead to nonuniform slip system activity and rotation within individual grains. This leads to the division of crystals into regions of differing lattice orientation. Bay et al. [2] observed that dislocation boundaries separate regions of a grain that have differing orientations resulting from nonuniform deformation in cold worked face-centered cubic metals. Furthermore, they found that the number of active slip systems in these regions was frequently less than that required to fulfill the Taylor criterion [3] (i.e., five active slip systems assuming uniform strain within each grain of a polycrystal). The same observation was made earlier by Fleischer [4], who counted the number of active slip systems in deformed polycrystalline brass in grains fully constrained by neighboring crystals. Fleischer found that approximately 40% of the grains had less than five active slip systems due to the nonuniform deformation within the grains. One reason for this is that at room temperature the strain rate sensitivity is very low leading to the deformation being mostly concentrated in localized regions of high slip, as shown in Peirce et al. [5]. Hughes et al. [6–8] studied the misorientations, i.e., the minimum relative angle that reflects crystal symmetry, that developed between the different regions of the grains and the dislocation boundaries that separated these regions as a result of nonuniform lattice rotation within the grains. They found that the misorientation distribution, for both incidental dislocation boundaries and geometrically necessary boundaries, scaled with the average misorientation angle, and the average misorientation angle, for both types of subgrain boundaries, scaled with the effective strain. The present work investigates, through simulation, the effect that additional active slip systems at elevated temperatures may have on intragranular misorientations.

The deformation microstructure (prior to heat treatments) that results at elevated temperatures and high strain rates, typical in industrial forming processes, shows some key differences to those observed in cold working. First, because the rate sensitivity is higher at elevated temperatures, the deformation is more diffused. Second, at room temperature, only the 12 octahedral slip systems \{111\}\{110\} are active, whereas at elevated temperatures (above a homologous temperature of 0.65), additional slip systems have been observed to be active (see, for example, Percheau and Driver [9], Kvapilova and Orlova [10], and Martin and Caillard...
sulting from the breakup of grains with unstable orientations, are sensitive over a wide range of mesh densities, and that with suff-

Other researchers have modeled polycrystals with discreti-

tions in a uniform grid are distorted to form Wigner–Seitz cells,

number of finite elements

utilizes a 17-grain, three-dimensional, representative polycrystal

sensitivity occurring at elevated temperatures have on the evolu-

structures is explicitly represented to investigate the response of a

In the present study, we apply an approach in which the grain

are imposed on the model polycrystal. In order to investigate a

larger number of orientations, simulations are performed consid-

ering three sets of representative orientations. Simulations allow-

ing for only the 12 octahedral {111}{110} slip systems are com-

pared to those that consider the octahedral slip systems as well as

slip on the additional 12 {110}{110} and {001}{110} slip systems.

The model is calibrated against a macroscale compression test,

and then the deformation history predicted by a macroscale sim-

ulation of the compression test is applied to the model polycrystals.

The predicted microstructures, considering both 12 and 24 slip

systems, are compared to the resulting grain structure from the

compression test as well as observations reported in the literature.

The experimental data are provided by Misiolek’s group at Lehigh

University, and similar compression test data on the same alloy

are described in Ref. [21].

2 Multiscale Polycrystal Model Formulation

2.1 Grain Scale Model. First, we consider the behavior on the

grain scale. Let us assume the motion \( x = \phi(\textbf{X}, t) \) and let

\( \mathbf{F}(\textbf{X}, t) = \nabla \phi(\textbf{X}, t) \) be the deformation gradient at the current time

and position \( 	extbf{X} \) in the reference configuration \( \Omega_0 \).

The crystal elastoviscoplastic formulation is fairly standard and follows that
given in Maniatty et al. [22] and is summarized here for complete-

ness. A multiplicative decomposition of the deformation gradient

into elastic and plastic parts, following the work of Lee [23], is

assumed, where the elastic deformation gradient \( \mathbf{F}^e \) can further be

decomposed into the lattice rotation \( \mathbf{R} \) and the elastic lattice

stretch \( \mathbf{U}^e \) by the right polar decomposition. We assume the plastic

deformation gradient \( \mathbf{P}^p \) is volume preserving and that crystallo-

graphic slip, accompanied by dislocation motion, is the only

mechanism for plastic deformation so that

\[
\dot{\mathbf{P}}^p = \mathbf{F}^e \dot{\mathbf{F}}^{-1} = \sum_{\alpha=1}^{N_s} \dot{\gamma}^\alpha \mathbf{P}^\alpha
\] (1)

where \( \dot{\mathbf{P}}^p \) denotes the plastic velocity gradient on the relaxed con-

figuration \( \tilde{\Omega} \), \( \dot{\gamma}^\alpha \) is the rate of shearing on slip system \( \alpha \), \( N_s \) de-

notes the total number of slip systems for the crystal, and \( \mathbf{P}^\alpha \)

\( = \mathbf{s}^\alpha \otimes \mathbf{m}^\alpha \) is the Schmid tensor. Here, \( \mathbf{s}^\alpha \) and \( \mathbf{m}^\alpha \) denote the slip
direction and slip plane normal, respectively. For the elastic be-
havior, a linear relationship, with cubic symmetry, is assumed
between the second Piola–Kirchhoff stress \( \mathbf{S} \) and the Green elastic
strain \( \mathbf{E}^e \) on the relaxed configuration \( \tilde{\Omega} \).

To complete the constitutive model, it remains to relate the stress
to the plastic deformation. This is accomplished by relating the
resolved shear stress on a slip system \( \tau^\alpha \) to the rate of shearing
on that slip system \( \dot{\gamma}^\alpha \) through a usual power law. Specifically,

\[
\dot{\gamma}^\alpha = \frac{\tau^\alpha}{G^\alpha} \left( \frac{\tau^\alpha}{G^\alpha} \right)^{m-1}
\] (2)

where \( \tau^\alpha \) is a reference shear rate, \( m \) is the strain rate sensitivity,
and \( G^\alpha \) is the resistance to slip on the \( \alpha \) slip system. The definition of
the resistance to plastic slip (flow stress) on the slip systems \( G^\alpha \)
is described next.

As mentioned in Sec. 1, in addition to the 12 compact slip
systems, slip is observed to occur on noncompact slip systems in
aluminum alloys at the temperatures of interest. In particular, the family of six \{001\}{110} slip systems and the family of six \{110\}{110} slip systems are also available, giving a total of \( N_s = 24 \) slip systems to consider. Based on the observations in Pero-
cheau and Driver [9], we assume the resistance to slip on all the
compact octahedral slip systems \{111\}{110} is the same, i.e., \( G^\alpha = g \), where \( \alpha \) is a compact slip system. Likewise, on each of the
noncompact slip system types, the flow stress is assumed to be the
same and related to that on the compact slip systems using the
ratios observed by Perocheau and Driver. It should be noted that
these ratios were measured in a different aluminum alloy system,
but in the absence of measurements for the alloy of interest, we take the same ratios. Specifically, for the \{011\}(110) slip systems, we take \( g^\alpha = 1.4 \)g and for the \{110\}(110) slip systems, we take \( g^\alpha = 0.9 \)g, where \( \alpha \) and \( \alpha_g \) refer to each of the different types of noncompact slip systems. Now, only an evolution equation for the reference flow stress \( g \) is required. At the elevated temperatures of interest, the saturation flow stress, due to the simultaneous dislocation entanglement hardening and softening by recovery, is quickly reached after little hardening. A simple Voce–Kocks [24,25] type of hardening model is used to capture this phenomenon:

\[
\begin{align*}
\dot{g} &= G_g \left( g_s - g \right) \dot{\gamma} \\
\dot{\gamma} &= \sum_{\alpha=1}^{24} |\dot{\gamma}| \\
g_s &= g_{s_0} \left( \frac{\dot{\gamma}}{\dot{\gamma}_0} \right)^{s} 
\end{align*}
\]

where \( g_s \) is the initial reference flow stress, \( g \) is the saturation reference flow stress, and \( g_{s_0}, G_g, \dot{\gamma}_0, \) and \( s \) are material parameters that need to be calibrated. The primary drawback of using this hardening model here is that it has no intrinsic length scale, so grain size and subgrain size effects are not captured. A model, such as that used in Beaudoin et al. [17], could capture such effects but is beyond the scope of the present work.

### 2.2 Macroscale-Grain Scale Link.

In addition to the constitutive behavior linking the deformation history to the stress and evolution of material state, the usual equations of equilibrium must be satisfied and appropriate boundary conditions must be prescribed that represent as closely as possible the physical situation. In the present study, deformation at the grain scale is linked to the macroscale by way of a deformation gradient history recorded at an interior point of a macroscale finite element simulation. Specifically, the deformation history is imposed on a representative polycrystal through the boundary conditions, see, for example, work by Hill [26] and Miehe [27].

Periodic boundary conditions are applied on the representative polycrystal model to reduce boundary condition effects, where the experiment sample is assumed to be made by stacking identical polycrystal representative volume elements (RVEs) periodically. Here, a locally periodic, first order method is used. First order, locally periodic methods are appropriate if the material of interest is at an interior location, the microstructure is uniformly random (i.e., no gradient in the microstructure), and the deformation gradient can be treated as uniform in an average sense on the fine scale. These conditions hold here. The macroscale deformation gradient \( \mathbf{F} \) is treated as being constant over the fine scale, and the position of a material point on the fine scale is assumed to be governed by this deformation gradient plus a fluctuation field such that the spatial coordinates of a particle at the grain scale are given by

\[
\mathbf{x} = \mathbf{F} \mathbf{X} + \mathbf{\bar{u}}
\]

where \( \mathbf{\bar{u}} \) denotes the fluctuation in the displacement field on the mesoscale. Equilibrium is then achieved by permitting a fluctuation field, which can be solved for using the finite element method. The fluctuation field is periodic and must match on opposite sides of the RVE. Because of triple periodicity at the eight corners of the RVE (assuming a cubic RVE), the fluctuation field at the corners must be equal and is set to zero.

The above kinematic description together with the equilibrium equation and elastoviscoplastic crystal constitutive equations defines a boundary value problem on the grain scale for the fluctuation field \( \mathbf{u} \). Assuming a prescribed macroscopic deformation gradient, the boundary value problem can be solved using the finite element method.

### 2.3 Numerical Implementation.

The constitutive equations described in Sec. 2.1 are integrated using the procedure described in Maniatty et al. [22] and implemented into a general purpose, parallelized finite element framework. A mixed formulation is used, where the increments of displacement fluctuation \( \mathbf{\bar{u}} \) and hydrostatic stress are interpolated separately to handle the nearly incompressible behavior at large strains. A consistent tangent is derived to ensure quadratic convergence in the solution of the resulting nonlinear system of equations. Details of the formulation can be found in Lu [28].

### 3 Polycrystal Model and Calibaration

The 6063 aluminum alloy considered here to calibrate and test the polycrystal model is from a section of a billet that was cast with the direct-chill process, and then homogenized at 560 °C for 4 h and cooled at a rate greater than 200 °C/h. For more details about the processing and composition, see Van Geertruyden et al. [21]. The processing resulted in a fairly equiaxed grain structure with no strong texturing. Figures 1 and 2 were produced using data from an EBSD scan of the material before deformation. Figure 1 shows a projection of both the grain structure and the texture, as the grain orientations are colored according to the inverse-pole-figure color map shown. The area scanned is 800 \( \times \) 1600 \( \mu \text{m}^2 \), and the step size is 4 \( \mu \text{m} \). Figure 2 depicts the texture through an inverse pole figure. Looking at the scale bar, the
texture appears fairly random with no strong texture components.

To mimic the observed initial grain structure in the polycrystal model, a three-dimensional polycrystal RVE was created using a modified Monte Carlo grain growth algorithm, following that presented in Radhakrishnan and Zacharia [29], and subsequent geometric modeling operations. This approach generates a model grain structure with equiaxed grains and a grain size distribution similar to that observed. The procedure for creating the representative polycrystal model is outlined below. A detailed treatment can be found in the thesis by Lu [28].

Creation of the polycrystal RVE involved three stages: grain growth using a modified Monte Carlo algorithm, extraction of topology information, and meshing. The grain growth algorithm operates on a regular structured grid, where each grid point is assigned an orientation, and groups of neighboring grid points with the same orientation define a grain. Here, a $50 \times 50 \times 50$ regular grid was used to define the RVE. In order to satisfy the periodicity assumption described in Sec. 2.2, periodic boundaries were created by linking opposite faces of the polycrystal RVE. A polycrystal with 17 grains (45 regions due to periodicity) was created. The Monte Carlo algorithm creates a voxel representation of the polycrystal RVE. To smooth the grain boundaries and allow for unstructured meshing for subsequent finite element analysis, a smooth geometric model was created from the voxel data. The smooth geometric model was generated by extracting topology information—vertices, edges, faces, and regions—from the voxel data. Finally, the polycrystal grain structure was meshed with 24,819 P2/P1 elements (tetrahedral elements, quadratic displacement interpolation, linear hydrostatic stress interpolation). While the grain structure is relatively small (only 17 grains), there is a good mesh resolution with on average about 1460 elements per grain to allow intragranular behavior to be studied. Grain orientations were assigned to the model polycrystal by sampling an experimentally obtained orientation distribution function. In order to partially compensate for the small number of grains in the polycrystal RVE, three sets of orientations were considered. The grain structure with the three sets of orientations is shown in Fig. 3 together with the inverse-pole-figure color map used to depict the orientations. An inverse pole figure showing the composite set of orientations considered is shown in Fig. 4. Since only 51 orientations are considered, the randomness of the initial texture is not well represented. While the sampling was random, a random sample that contained a preexisting (100) texture component was selected since the work of Stout et al. [14] observed that this texture component was stabilized at elevated temperatures.

The calibration work focused on determining the parameters in the constitutive model (namely, $\gamma_0$, $m$, $G_0$, $\bar{G}_0$, $\bar{G}_s$, $\bar{\gamma}_s$, and $\omega$ in Eqs. (2)–(5)) using available experimental results, which include a compression test performed at Lehigh by Misiolek’s group and information in the literature. To narrow the scope of this work, and also due to the limited available data, only isothermal conditions were considered at a temperature of 482°C, which is the test temperature of the compression test performed at Lehigh. The compression test was performed on a 15 mm long specimen with a 10 mm diameter that was produced as described at the beginning of Sec. 3. The test consisted of compression at 1 mm/s for 10 s to a final height of 5 mm (33% of the original height or a 67% reduction).

The calibration work consisted of two steps. First, an isotropic hardening model, analogous to that in Eqs. (3) and (4), was calibrated against the available stress-strain data from both literature and compression tests. Then, by matching the stress-strain curve of a polycrystal model under uniaxial tensile loading to the stress-strain curve generated by the calibrated isotropic hardening model, the polycrystal hardening model parameter values were determined. Different polycrystal models, made up of cube shaped grains, were used in producing the uniaxial tension stress-strain curve to ensure that the parameter values are model independent.

![Fig. 2 Inverse pole figure showing texture for cast and homogenized 6063 aluminum alloy prior to deformation. The image is with respect to the [010] sample direction, where [010] is the loading direction.](image1)

![Fig. 3 Initial model grain structures. (a), (b), and (c) depict the three sets of orientations considered. The inverse-pole-figure color map is shown on the right and is with respect to the [010] sample direction (compression axis). The same color map is used for all simulation results.](image2)

![Fig. 4 Inverse pole figure made prior to deformation showing the composite set orientations used in the simulations. The [010] direction corresponds to the loading axis in the compression test.](image3)
The final model parameters selected are given in Table 1. The elastic parameters for aluminum at 482°C, $C_{11}=93.1$ GPa, $C_{12}=61.7$ GPa, and $C_{44}=21.0$ GPa, were also used.

### Table 1 Calibrated hardening model parameters

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\dot{\gamma}_0$</td>
<td>3.0 s$^{-1}$</td>
</tr>
<tr>
<td>$\omega$</td>
<td>0.0289</td>
</tr>
<tr>
<td>$\dot{\gamma}_s$</td>
<td>$6.8634 \times 10^{-8}$ s$^{-1}$</td>
</tr>
<tr>
<td>$m$</td>
<td>0.11</td>
</tr>
<tr>
<td>$G_o$</td>
<td>91.23 MPa</td>
</tr>
<tr>
<td>$g_o$</td>
<td>16.63 MPa</td>
</tr>
<tr>
<td>$g_{so}$</td>
<td>16.51 MPa</td>
</tr>
</tbody>
</table>

The simulations, representative polycrystal models with the three sets of grain orientations considered were compressed to 60% of their original height using the deformation gradient history obtained from the macroscale compression simulation near the center of the specimen. This is not the total compression that was attained in the experiment and in the macroscale model, which was 33% of the initial specimen height, because the finite elements become too distorted to continue. For each grain structure, in order to see the effect of the additional sets of slip systems that become active at elevated temperatures, we ran cases with only the 12 compact, octahedral slip systems available as well as cases with all 24 slip systems available, using the same set of material parameters. Figure 7 shows the resulting inverse pole figures for the deformed grain structures comparing the cases with 12 slip systems with those allowing for 24 slip systems. Note that these inverse pole figures are with respect to the compression axis ([010]). We can see in Fig. 7(a) that when only 12 slip systems are considered the usual (110) texture component dominates, as expected. In Fig. 7(b), we see that the texture predicted with 24 slip systems still shows a dominant (110) texture, but a (100) component is also present. This is consistent with the observations in the color variation within the grains. Figure 6 depicts the texture through inverse pole figures. Here, we also show the [010] inverse pole figure, which shows the texture developed along the compression axis. The highest texture components are (110) and (100), which is consistent with other experimental observations for aluminum alloys at elevated temperatures, see Stout et al. [14].

## 4 Numerical Results and Discussion

Before presenting the simulation results, we first present the experimental observations. The sample compression axis is the $y$ axis. Figure 5 shows the resulting experimental grain structure near the center of the specimen and some of the local texture. Notice that there is a preponderance of red and purple and the color is nonuniform within the grains. Comparing Fig. 5 to the initial grain structure shown in Fig. 1, we see that the grains are more elongated and tend to have an orientation with the (111) or (100) direction along the transverse direction ([001]), and that the orientations within the grains are no longer uniform as depicted by the color variation within the grains. Figure 6 depicts the texture through inverse pole figures. Here, we also show the [010] inverse pole figure, which shows the texture developed along the compression axis. The highest texture components are (110) and (100), which is consistent with other experimental observations for aluminum alloys at elevated temperatures, see Stout et al. [14].
Stout et al. [14], who attributed the strong (100) final texture to a fairly strong initial cube texture that is stabilized by slip on the {110} planes.

Figures 8 and 9 show the resulting compressed grain structures allowing for 12 and 24 slip systems, respectively. The same [010] (compression axis) inverse-pole-figure color map used in Fig. 3 is used here to show the orientations. The periodicity of the deformed structures that was enforced through the boundary conditions is evident. Notice that the grains no longer have a uniform orientation, but the case allowing for 24 slip systems shows less orientation variability within the grains. It is also interesting to note that the deformation with 24 slip systems is considerably more uniform, which is not surprising given the additional deformation paths available. One way to quantify the nonuniformity of the deformation is to compute the average fluctuation in the displacement field from that of a uniform deformation. The average fluctuation in the displacement field for the 12 slip system cases is 2.4 times higher than that for the 24 slip system cases.

To make a more direct comparison to the experiment, we made a cut through the grain structures on an x-y plane in the interior of the grain structure (y is the compression axis) and show the before and after grain structures in Figs. 10–12. A rectangular region from within the cross section was used due to the restrictions of the imaging software. Figure 10 shows the original cross section considered with the initial orientations using the [010] inverse-pole-figure color map. Comparing Figs. 11 and 12, we see that the orientation variation within the grains appears to be greater for the 12 slip system case. To better show the orientation variation within the grains (misorientation), Fig. 13 shows the kernel average misorientation on a slice of the second orientation set comparing the case with 12 slip systems to that with 24 slip systems. The kernel average misorientation for a given point is the average misorientation between the point and all of its neighbors excluding misorientations greater than some prescribed value, 15 deg in this case, so as not to consider grain boundaries. In generating the results in Fig. 13, the kernel average misorientation was computed using the third nearest neighbors. We can see in these images that the case with 24 slip systems predicts lower typical misorientations over much of the cross section. These results are consistent with the observations described in Ringeval et al. [13], who observed that at elevated temperatures, the orientation distribution within grains is quite uniform. In addition, it is interesting to

![Fig. 7](image-url) [010] (compression axis) inverse pole figures showing predicted resulting textures where the results from all the three orientation sets are combined: (a) 12 slip system cases and (b) 24 slip system cases.

![Fig. 8](image-url) Polycrystals compressed to 60% of their initial height with 12 slip systems available. (a), (b), and (c) are the three orientation sets considered. The inverse-pole-figure color map is with respect to the [010] sample compression axis.

![Fig. 9](image-url) Polycrystals compressed to 60% of their initial height with 24 slip systems available. (a), (b), and (c) are the three orientation sets considered. The inverse-pole-figure color map is with respect to the [010] sample compression axis.
investigate the distribution of misorientations. Following the procedure used in Dawson et al. [18], where the misorientations are computed across element boundaries, we compute the misorientation distributions for a single grain for both the 12 and 24 slip system cases. The result is shown in Fig. 14. Other grains were also investigated, with similar results, so only the single case is shown. Interestingly, the scaled distributions are very similar to each other and to those computed in Dawson et al. and those observed experimentally in Hughes et al. [6]. However, the average misorientations, used to scale these distributions, are quite different, with the average being 3.3 deg for the 12 slip system case and 1.84 deg for the 24 slip system case.

Finally, we investigate the slip system activity. Figure 15 shows the probability distribution for the number of active slip systems based on the model predictions for the 12 and 24 slip system cases. To define whether a slip system is active, we set a slip rate cutoff of $10^{-5}$ s$^{-1}$; if the slip rate is higher than this cutoff, the slip...
localized at cold temperatures, which is correlated with low rate sensitivity; at elevated temperatures, with a higher rate sensitivity, the deformation is more diffused (see, for example, Peirce et al. [5]). When the deformation is nonuniform within the grains, the requirement of at least five active slip systems for constrained grains is no longer necessary and, in fact, is frequently not observed. Furthermore, as observed in simulations by Maniatti and Yu [30], the number of active slip systems is also reduced by elastic effects, which continue to cause transients as the grains reorient and the stress state traverses the flow surface. However, at elevated temperatures, especially with additional slip systems available, the deformation within the grains is not as localized and is more uniform with higher numbers of active slip systems.

5 Conclusions
An overall framework for modeling representative polycrystal RVEs was presented with the model being specialized to consider an aluminum alloy at elevated temperatures. Specifically, a model that allows for both the usual octahedral \{111\}(110) slip systems and the nonoctahedral \{110\}(110) and \{100\}(110) slip systems, which are active in aluminum alloys at elevated temperatures, is described. Simulation results considering only the 12 octahedral slip systems are compared to those allowing for all 24 slip systems, and the simulations are compared to experimentally observed microstructures resulting from a compression test conducted on a 6063 aluminum alloy. The simulations predict a more uniform deformation, and as a result, more uniform orientations within the grains, when 24 slip systems are available. This is consistent with the experimental observations reported in Ringeval et al. [13] and is not surprising given the additional deformation paths available. The higher strain rate sensitivity at elevated temperatures also contributes to the uniformity of the deformation. The texture predicted for the 24 slip system case is qualitatively similar to the experimental texture with both \{110\} and \{100\} texture components observed. While the average intragranular misorientation in the 12 slip system case was higher than that for the 24 slip system case, the scaled misorientation distributions for each case are similar. The number of active slip systems throughout the polycrystals was also recorded at maximum deformation allowing for the probability of a given number of active slip systems to be predicted. When 24 slip systems are allowed, the number of active slip systems (mostly in the range of 10–14) is usually larger than the number of slip systems when only 12 slip systems are available (typically 6 or 8).

The primary limitations of this work are the small grain structure considered, which is not sufficient to be truly representative, and the small number of orientations considered, which do not adequately represent the initial texture. The maximum deformation is also limited due to the element distortion. Modeling larger polycrystal structures with more grains and considering more set of orientations are topics of future work. Work is also being done to allow for remeshing so that larger deformations may be considered in the future.

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