Influence of Geometric Dynamic Recrystallization in the Dead Metal Zone on Peripheral Coarse Grain Defect Formation in Indirect Extrusion

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Careful metallurgical analysis of partially indirectly extruded 6061 billets has revealed a region of very fine, equiaxed, high-angle boundary (HAB) grains in the dead-metal zone (DMZ), "leaking" into the region of the extrudate which subsequently develops into the peripheral coarse grain (PCG) defect. The fact that the DMZ may not be completely "dead", and that a microstructural evolutionary phenomenon is creating very fine HAB grains in this region, which then develop into the PCG defect, inspired a more thorough experimental investigation and microstructural analysis of the indirect extrusion process, and of microstructure evolution of the defect-forming region. Finite Element Modeling (FEM) analyses indicate that indeed the DMZ *does* undergo significant deformation during indirect extrusion, rather than remaining completely static at the die face, thereby suggesting that there is sufficient deformation in this region for Geometric Dynamic Recrystallization (GDRX) to occur. As existing quantitative models for GDRX tend to underpredict the fraction of HABs present, a more detailed mesoscale model of the GDRX phenomenon is presented, with particular attention to the 3-D geometry of grains, their ratio of low-angle boundaries to high-angle boundaries, the effect of grain shape change (GSC) during deformation, and other phenomena. Experimental results, FEM simulations, and EBSD analyses are presented.

MOTIVATION

With increased competition and rising demands on the quality and cost of extruded products, there is a strong need for "value-added" engineering in the aluminum extrusion industry. Improved extrudate properties, higher yield per billet, and fewer defects are all possible methods to add additional value to the process. With improved understanding of the underlying science of the processing→ microstructure→ properties cycle, and improved capability to model these phenomena, new tools to permit optimizing the extrusion process without costly trial and error are increasingly available.

BACKGROUND - PERIPHERAL COARSE GRAIN DEFECT

One common defect in the extrusion process is the Peripheral Coarse Grain (PCG) defect. This consists of a ring of abnormally large grains at the periphery of an extrudate, which can vary in depth from a mere surface effect to one which consumes nearly the entire volume of the product. The PCG region is typically of inferior mechanical properties, and is often machined away, adding cost to the final product. The PCG defect forms *after* the extrudate has left the die orifice, suggesting either a post-dynamic recrystallization phenomenon (e.g. metadynamic or static recrystallization), or abnormal grain growth (AGG), or some combination of the two. While characterizing this defect, Van Geertruyden[1] observed evidence of abnormally large grains apparently slipping out of the so-called dead-metal zone (DMZ) of the billet and into the periphery of the extrudate (Figure 1). With the possibility that the PCG defect originates from grains slipping out of the DMZ thus raised, Van Geertruyden more fully investigated the PCG region, and Van Geertruyden and Browne[2] more fully characterized the DMZ region via optical and Electron Backscatter Diffraction (EBSD) techniques. The observation of fine, equiaxed, high-angle boundary grains, similar to recrystallized grains, in the supposedly static DMZ region, coupled with the observation that they may be slipping into the PCG region of the extrudate, encouraged further examination of the material flow and microstructural phenomena within the DMZ and into the PCG – particularly with regard to the typical "textbook" consideration that the DMZ does not deform much, and hence should not undergo significant microstructural evolution, let alone very fine refinement and possible subsequent evolution into the PCG. To investigate this possibility, Bandar[3] performed Finite Element Modeling (FEM) and microstructural evolution simulations of the material in the DMZ, and determined that rather than experiencing little or no deformation in the traditionally "static" area of the DMZ, the material in this portion of the billet, in fact, undergoes significant hot deformation. Returning to Valberg's hallmark visioplasticity studies of direct and indirect aluminum extrusion[4] (Figure 2), it is apparent that during *direct* extrusion, the DMZ does in fact appear to be quite "dead" – the original grid lines and/or material slices at the die face persist throughout the extrusion process, with little or no displacement; the material is effectively fixed to the die face due to sticking friction. However, careful observation of *indirect* visioplastic results indicates that the material at the die face is not quite static, and in fact is slipping into the periphery of the extrudate.

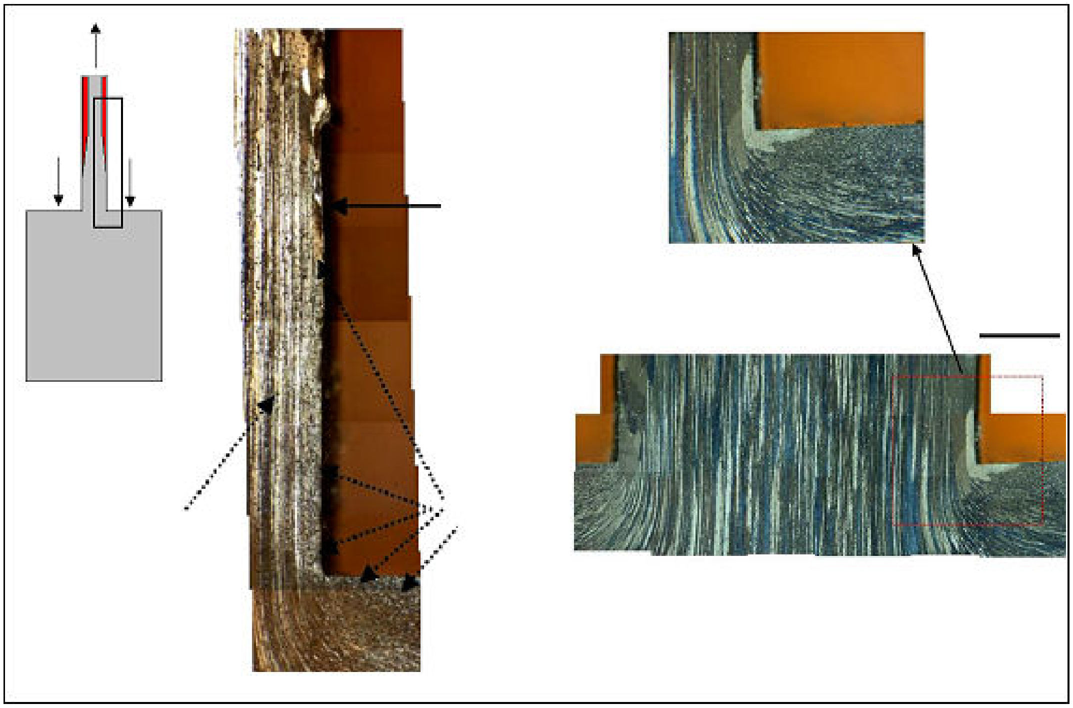

Figure 1. Evidence of fine grains in the DMZ region of indirect aluminum extrusion flowing into the extrudate, and subsequently becoming the Peripheral Coarse Grain (PCG) defect (W. Van Geertruyden[1]).

In this regard, the material flow at the die face of an indirect extrusion may be considered similar to the material flow of the "piping" defect seen at the back of a billet during direct extrusion; as the indirect extrusion die advances into the billet, it pushes material forward, and the material slides across the die face towards the center of the billet and out through the orifice into the extrudate. For the direct process, as the ram proceeds through the billet, the material slides across the ram face and ultimately gets drawn up into the center of the billet, leading to the "piping" or "coring" defect. Given that, for the indirect process, this material then occupies the periphery of the extrudate, it raises the obvious consideration that the grains of this region subsequently evolve into the PCG. The question remains – what is unusual about these grains that later leads to a very coarse microstructure, that is distinct from the bulk of the extruded grains, which remain long and fibrous in morphology? The following FEM/microstructural analysis attempts to address this concern.

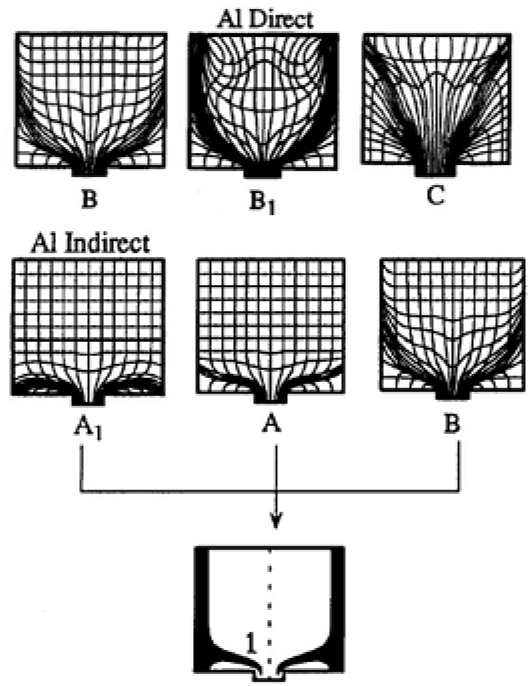

Figure 2. Valberg's characterization of material flow during extrusion[4].

FINITE ELEMENT MODELING AND MICROSTRUCTURE EVOLUTION

A review of the literature revealed no satisfactory explanation for the formation of this defect from the billet into the extrudate. Thus, to analyze the microstructural evolution of grains flowing from the billet into the DMZ, and then into the extrudate, and finally possibly becoming the PCG, a new microstructure evolution model was written, informed by state variables (strain, strain rate, and temperature) determined via FEM.

Finite Element Modeling of Indirect Extrusion

Multiple extrusion simulations were performed, using the FEM code DEFORM™. The alloy studied was 6061 aluminum, for which thermomechanical material data exists in DEFORM™ (flow stress, thermal conductivity, etc). It should be noted that the "default" friction condition for hot forming of aluminum in DEFORM™ is given as a shear factor of 0.4. Whereas this may be reasonable for open-die forging, this is inconsistent with the "sticking-condition" common in aluminum extrusion, and was changed to 1.0; this was the only parameter changed in the simulation from the default values.

In order to determine the accuracy of the FEM predictions, the material flow results were compared with visioplastic results. Although the experimental observations of grains slipping from the DMZ into the extrudate were made on billets which had undergone indirect extrusion, the most readily available sequence of visioplastic images of extrusion in the published literature were for a direct extrusion process, by Valberg[5]. Thus, a direct extrusion FEM simulation was performed, and "flownets" representing the material flow pattern were compared side-by-side with visioplastic images of a real extruded billet (Figure 3). Specific features on the visioplastic images related to the material in the DMZ were measured (Figure 4), and quantitatively compared to the equivalent material flow in the DEFORM simulations (Figure 5). The material flow predictions calculated by DEFORM were deemed reasonably accurate, and it was assumed that similar material flow predictions for an indirect extrusion simulation (Figure 6) would also be reasonably accurate. Additionally, the predicted temperature at the die face was compared with thermocouple measurements of the experimental extrusion, and also deemed sufficiently accurate for this study (at the end of the indirect extrusion experiments, the die face temperature was 425°C; FEM predicted 427°C).

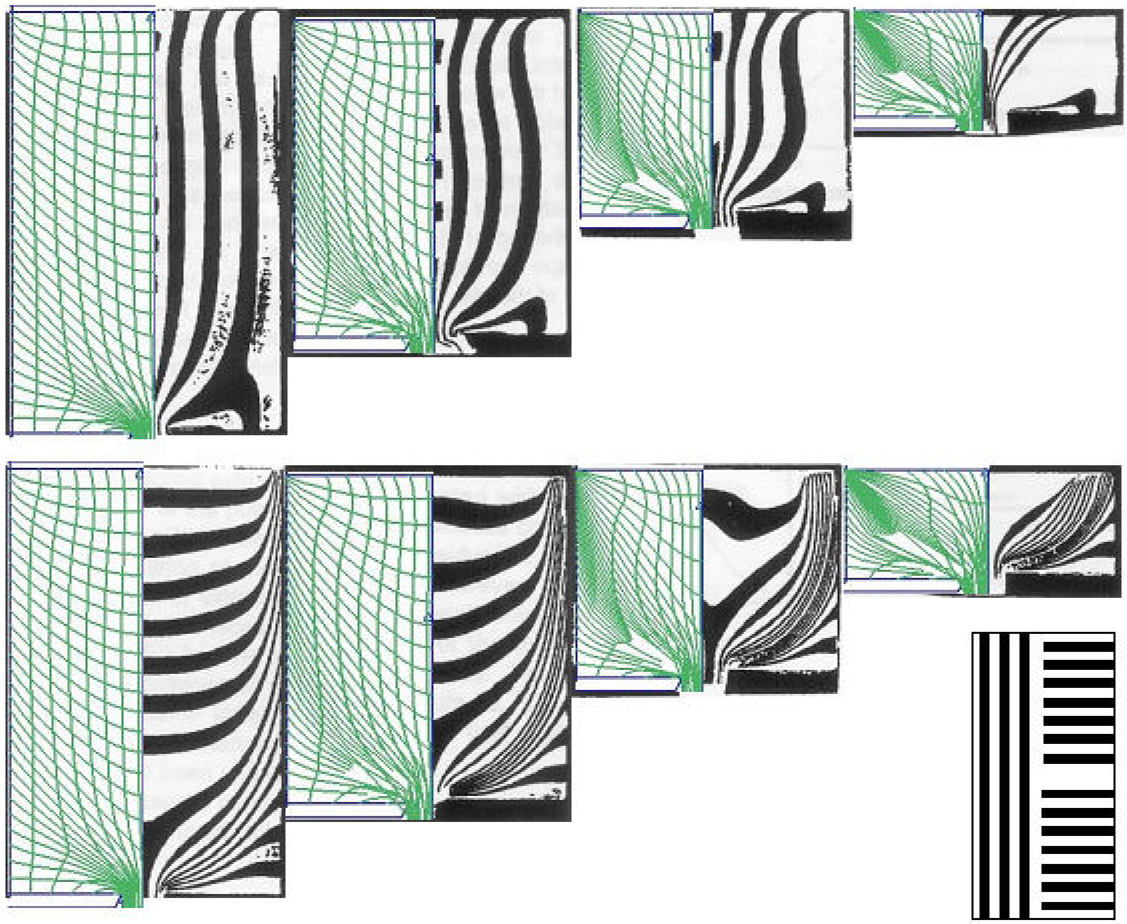

Figure 3. Comparison between FEM-predicted material flow and Valberg's visioplastic analyses of aluminum extrusion[5].

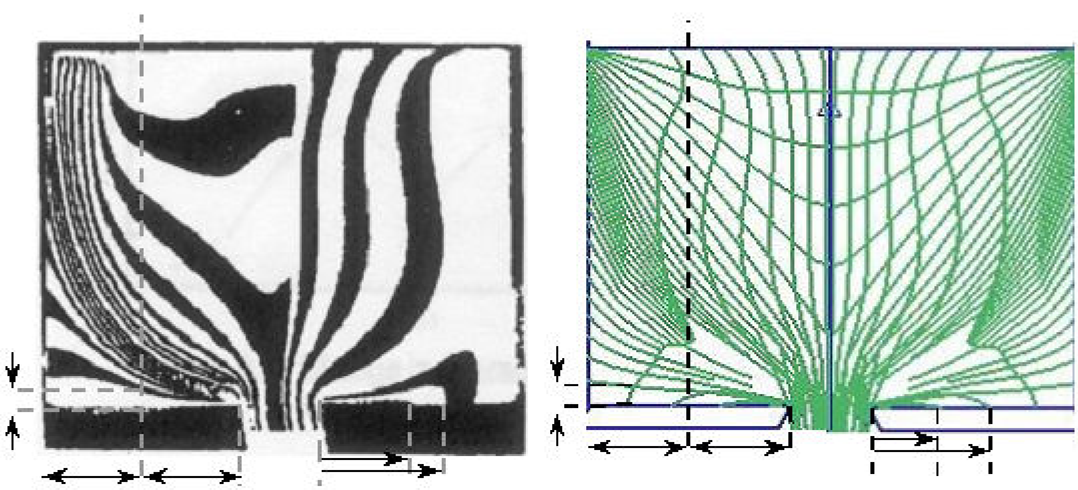

Figure 4. Metric points measured for evaluation of accuracy between FEM-predicted material flow and Valberg's visioplastic analyses[5].

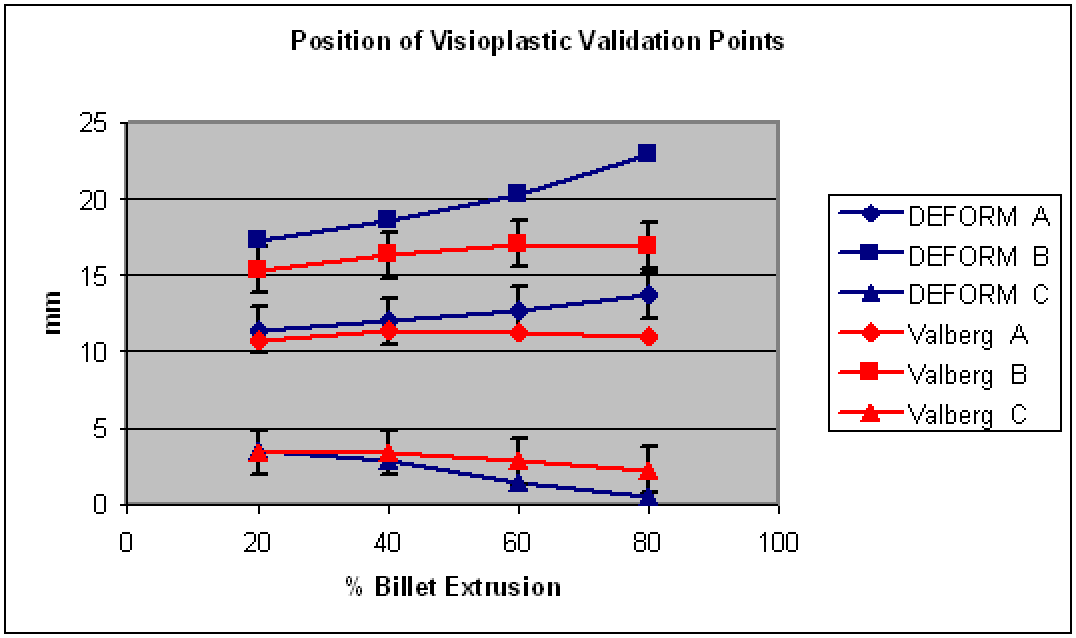

Figure 5. Quantitative comparison of the metrics used to evaluate the FEM-predicted material flow and the visioplastic analyses.

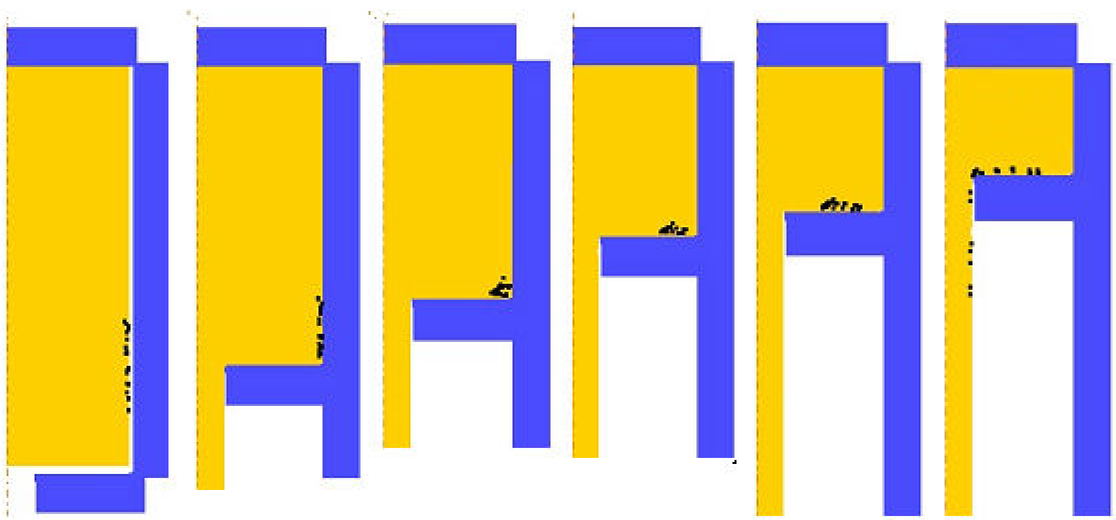

Figure 6. FEM point-tracking of material flowing from the edge of the billet through the DMZ and into the periphery of the extrudate, eventually becoming the PCG defect.

Microstructure Evolution Modeling of Indirect Extrusion

Typical empirical microstructure models, such as the Johnson-Mehl-Avrami-Kolmogorov model (i.e. JMAK, or simply, Avrami model) are generally not capable of capturing such inhomogeneous microstructural effects as the PCG defect observed in this study. Such empirical models characterize the response of a microstructure (e.g. grain refinement via recrystallization) to field variables such as strain, strain rate, temperature, and time, by performing a series of calibration tests (such as compression tests on cylindrical or double-cone samples), and fitting nucleation and growth constants to an exponential curve. In this manner, predictions of recrystallized volume fraction and subsequent average grain size may be made. However, these predictions are not phenomenological; they are only valid if the thermomechanical processing envelope of the calibration tests encompass that of the actual forming process, and if the phenomena characterized (such as DRX) are the overwhelmingly significant factors in the microstructural evolution. If the forming process enters a different thermomechanical regime, or experiences phenomena other than that of the calibration tests, then the predictions may not be accurate. Since the unusual microstructural evolution observed in the DMZ of indirectly extruded aluminum has not been well characterized, a more phenomenologically based microstructure modeling approach was employed.

Mesoscale modeling

One approach to more phenomenologically-based microstructure modeling is to characterize and catalog the microstructure features themselves, and update them with physical mechanisms, rather than to empirically curve fit a JMAK-style exponential function to them. This type of "microstructural bookkeepping" permits analysis of the impact of various different mechanisms, and allows evaluation of the significance of different phenomena. This type of modeling has been published previously for aluminum by Gourdet and Montheillet[6], and by Nes[7], and for other FCC metals such as nickel-base superalloys by J. P. Thomas[8]. In these types of models, microstructural features such as grain boundary density, grain boundary angular misorientation distributions, and dislocation density are evolved as functions of mechanistically-based equations. Hardening and recovery models such as Kocks-Mecking or Laasraoui-Jonas models are employed to evolve the dislocation density as a function of strain, strain rate, and temperature. Nucleation and growth mechanisms to model recrystallization, either as functions of critical dislocation density or critical grain boundary misorientations are employed to monitor grain refinement. Grain growth is modeled via grain boundary velocities as functions of time, temperature, and misorientation. Flow stress can be computed as a function of all of these features. Specific details of this model will be published later.

Dynamic Recrystallization in Aluminum

One of the challenges of this particular microstructural analysis is how to account for the apparent dynamic refinement of the grains in the DMZ. Aluminum, as a high-stacking fault energy material, does not undergo dynamic recrystallization in the traditional sense of discontinuously-nucleated new, dislocation-free grains, as other materials such as nickel, steels, titanium, and magnesium do[9]. Rather, two other potential mechanisms for dynamic recrystallization – geometric (GDRX) and continuous (CDRX) may be the cause of this apparent recrystallization[10]. However, there are not many published quantitative studies on the analysis of GDRX or CDRX in aluminum, and none regarding their effect within the DMZ. Furthermore, disagreement exists as to the exact quantitative impact of these models. Comparisons between the initial grain boundary densities and predicted final grain boundaries (after deformation and related microstructural evolution) do not accurately reflect experimental observation – individually, CDRX and GDRX models tend to underpredict the quantity of experimentally observed grain boundaries[11]. However, since it is possible, and in fact probable, that both of these dynamic mechanisms occur during deformation, it was decided to perform a microstructural evolution analysis where the effect of both phenomena were simultaneously modeled. Comparison of the predicted grain boundary densities would then be compared with the observed values, and if close, perhaps this combined effect could explain the formation of the fine, equiaxed grains in the DMZ, and perhaps subsequent effect on the PCG in the extrudate.

Combined CDRX and GDRX Mesoscale Model

A quantitative model of CDRX evolution in aluminum has been published by Gourdet and Montheillet[6]; however, no phenomenological quantitative model of GDRX evolution in aluminum has been published which could be combined with the Gourdet model, in terms of "keeping track" of the same microstructural variables as subgrains, misorientation distribution, and dislocation density. Thus, a new, phenomenological, quantitative model of GDRX was written, which could be integrated with the Gourdet CDRX model, in an attempt to evaluate their combined impact on grain refinement in the DMZ. Furthermore, an attempt was made to improve upon the accuracy of existing GDRX models, by eliminating some of the assumptions in other models, and calculating explicitly the grain boundary densities evolved as a result of GDRX. A brief description of the model follows; a more rigorous description will be published elsewhere.

Continuous Dynamic Recrystallization (CDRX). The CDRX component of the model is nearly identical to Gourdet's model. Readers are directed to the description of Gourdet's model for details; a brief description follows. This model computes subgrain formation rate and size as a function of dislocation hardening and recovery, and the evolution of these subgrains into grains due to dislocation absorption. As a material deforms, dislocation density increases due to Orowan looping, among other effects. Dislocation density decreases as a result of dynamic recovery. On balance, dislocation density increases, and these "excess" free dislocations eventually become tangled in cell walls, which "polygonize" into subgrains – grain boundaries of low misorientation angle. In this manner, an initially undeformed, dislocation-free grain forms subgrains within its boundaries. As these subgrains continue to absorb more dislocations, their relative misorientation increases, until they exceed a critical misorientation angle around 15-20 degrees. At this point they become high-angle boundaries, and thus produce a recrystallized microstructure.

Geometric Dynamic Recrystallization (GDRX). GDRX assumes that during deformation, grains form internal subgrains, and as the grains continue to deform, eventually they stretch or thin to the dimension of these subgrains. When this occurs, the subgrains within "break out" of their parent grains, and come into contact with other subgrains. This generally results in a new, high-angle grain boundary, resulting in a recrystallized-like microstructure.

Most GDRX models use a simplistic assumption of subgrain size, such as that given by Humphreys[12]:

$$\sigma / G \cdot \delta / b = K$$

where
σ =stress
G = shear modulus
δ = subgrain diameter
b = Burgers vector
K = a constant for aluminum (approximately 10 for FCC metals)

Additionally, most published GDRX models use a fairly simplistic geometric assumption of the critical strain at which subgrains "break-out" of their original grain. They assume a grain is a simple geometry such as a sphere or a cube, and when that shape becomes deformed to the same dimension of the subgrains (also taken to be spheres or cubes), it is assumed that all of the subgrains within "break out" simultaneously, and become new grains.

However, in reality grains/subgrains are not such simple shapes. If one were to represent a volume of microstructure by a collection of spherical or cubic grains, the volume may match, but the surface area would not. The difference in the surface area computation alone is on the order of 30-40%[3], and given that important microstructural reactions such as the nucleation of new grains, and the evolution of grain boundary misorientation, occur at high energy locations such as grain boundaries, these errors will rapidly become compounded.

Rather than simple geometric shapes, grains are complex, 3-D space-filling objects. Additionally, as a grain becomes deformed, the thickness at sharply faceted corners of the grain becomes thinner faster than the bulk of a grain; thus, subgrains at those regions "break-out" sooner, at a lower strain than the rest of the grain. Thus, if a grain were to be more accurately represented via a space-filling, 3D geometry, and if the deformation of that geometry were more accurately modeled, then presumably a more accurate depiction of the kinetics of GDRX may be modeled. Furthermore, if subgrain formation rate and size were modeled more phenomenologically, for example via Gourdet's CDRX model, rather than the Humphreys relation, then it is possible to mechanistically link GDRX with CDRX, and model these two phenomena simultaneously. The following model attempts to do so.

In order to more realistically represent the shape of grains and subgrains in this GDRX model, the shape of a grain/subgrain is assumed to be a tetrakaidecahedron, or more specifically, a truncated octahedron (TO). This is a 14-sided space-filling figure, used often to model the stereology of grains, compacted powders, and more[12]. The shape of the subgrains is also given by a truncated octahedron

(Figure 7). By using a space-filling shape such as a TO, it is possible to more closely match the volume and the surface area of real grains within a microstructure, at the expense of a more complicated analysis of the evolution of its shape. Given a certain strain, it is relatively straightforward to compute a new, updated shape and thus surface area (i.e. grain boundary area) of a sphere or cube. However, given a TO, it is more difficult to predict the new shape and surface area as it deforms. Thus for this GDRX model, the geometry of a TO, as it deforms in either tension, torsion, or compression, is updated by keeping track of the evolution of its vertices via a specially written Matlab™ code[3] (Figures 8, 9).

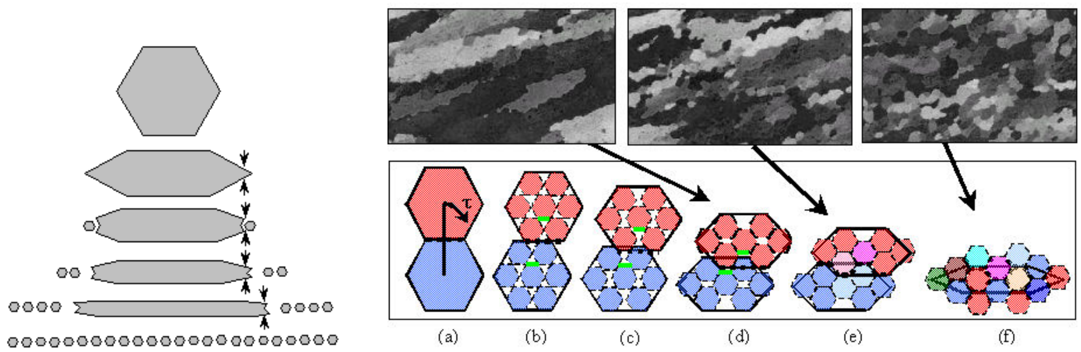

Figure 7. Left - simple depiction of the geometric dynamic recrystallization (GDRX) phenomenon as modeled by subgrains "breaking out" of a parent grain during deformation. Right, top - micrographs of GDRX grains presented by Pettersen et al[10]; Right, bottom - comparison to simple GDRX model of hexagonal grains and subgrains.

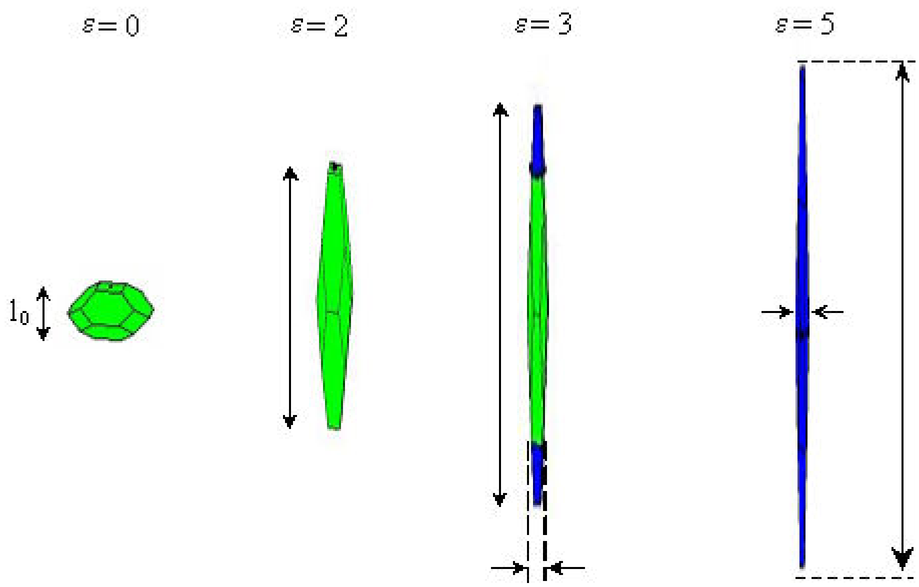

Figure 8. Left - computer representation of a parent grain represented by a space-filling truncated octahedron. Left, center - as the grain stretches in tension, it remains constant volume. Right, center - as it deforms further, the ends thin below the diameter of subgrains within the parent grain (these regions represented by darker/blue color). Right - after a certain strain, the entire grain is stretched below the diameter of subgrains, and it is entirely geometrically dynamically recrystallized.

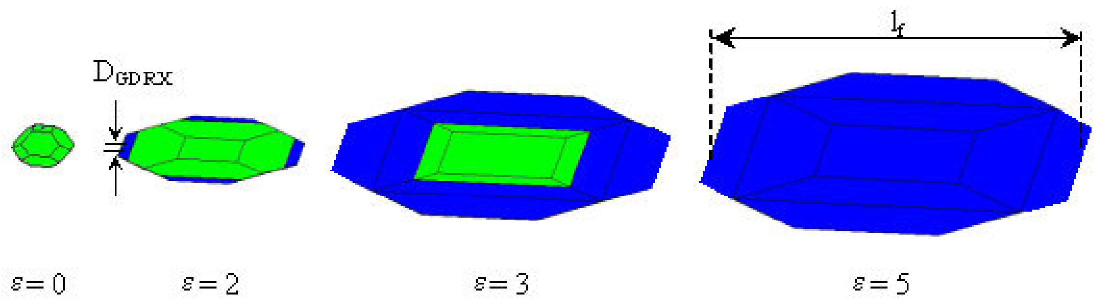

Figure 9. Left - computer representation of a parent grain represented by a space-filling truncated octahedron. Left, center - as the grain flattens in compression, it remains constant volume. Right, center - as it deforms further, the edges thin below the diameter of subgrains within the parent grain (these regions represented by darker/blue color). Right - after a certain strain, the entire grain is flattened below the diameter of subgrains, and it is entirely geometrically dynamically recrystallized.

EXPERIMENT

Aluminum 6061 with a low chromium content was deformed in torsion experiments, and Al 6061 with both low and high chromium contents (0.17 and 0.3 % respectively) were deformed via indirect extrusion experiments. The high chromium content helped reduce recrystallization via Zener pinning at chromium containing dispersoids, although the existence of large, recrystallized grains in the PCG of the high chromium samples indicate that clearly this pinning is insufficient to control the final grain size, and that some other modification to the TMP parameters must be made to improve product yield.

The torsion experiments involved deformation at strain rates of 15 1/s and 30 1/s, and at temperatures of 400 deg C and 482 deg C[1]. The initial grain size was approximately 125 microns. Micrographs of the samples were taken at radii which represented strains of approximately 1 and 3.5 (as computed via FEM analyses after the tests). After deformation, the samples were quenched and brought to less than 100 deg C within 1 second, and to room temperature within 2-3 seconds. Stepped annealing experiments on Al 6061 suggest that this was a sufficient quench rate to capture the deformed microstructure prior to any metadynamic/static evolution[13].

The indirect extrusion experiments involved cylindrical billets of low and high chromium-content Al 6061, deformed at a ram speed of 2.6 mm/s, at 482 deg C, into a rod of extrusion ratio 20. The initial grain size was approximately 100 microns. Since the TMP parameters varied as the material flowed through the billet, strain rates and temperatures were calculated via FEM and exported to the microstructure model. The quench rates were slower for these billets than for the torsion tests, as it was necessary to remove the billets from the extrusion press prior to quenching; however, the billets were quenched to room temperature within 30 seconds of the tests[2].

RESULTS

To evaluate the microstructure evolution model, the results of the torsion test were compared with the model's predictions. A torsion test was selected because a range of readily-calculable strains and strain rates are present throughout the cross-section, and thus the fractions of grains / subgrains can be measured via EBSD as a function of temperature, strain, and strain rate with a single test. If these predictions proved to be reasonably accurate, then the model would be applied to the more complex study of indirect extrusion.

The principal microstructure features evolved during deformation and recrystallization were (1) the volume and surface area of original grains; (2) the volume and surface area of subgrains; (3) the misorientation distribution of grains and subgrains; and (4) the dislocation density of deformed grains. The regions which were experimentally observed and characterized in the torsion sample were tracked in an FEM simulation. Each point tracked during microstructure modeling was given an initial grain size of 124 μm (taken from analysis of the real material prior to the torsion test), a "high angle boundary" distribution, and a low (0.01 $\mu m^2/\mu m^3$) dislocation density, typical for annealed metals[6]. As each point experiences strain in the FEM simulation, it is assumed that the TO grains modeled at that point strain to the same value. Thus, a grain “elongates” (if in tension or torsion) or “compresses” (if in compression) during deformation. The change in shape of the grain was computed via the Matlab™ code which updates the shape of its representative TO.

One surprising phenomenon determined during this computation was that during grain shape change (GSC), the surface area of a grain increases very significantly. It is assumed that the volume of an original “parent” grain doesn’t change, until its subgrains begin to break out due to GDRX. Since this and other GDRX/CDRX models tend to compute the fraction of subgrain boundaries present vs. total grain boundaries, it is important to accurately keep track of both during this type of computation. A deviation either in the density of subgrains formed, or in the density of existing grain boundaries stretched, will change this ratio, and comparison to EBSD-measured subgrain/grain ratios will be inaccurate. Thus, accurate modeling of the increase in grain boundary area due to GSC, as modeled via the TO model, is crucial.

Additionally, the kinetics of GDRX occur over a range of strains, as portions of the TO become thinned to the dimensions of the subgrains within. This is distinct from other published GDRX models, which compute a single critical strain, and then assume all of the subgrains break out at that strain. Results of the ratio of subgrains to grain boundaries as a function exclusively of grain shape change, of grain shape change + GDRX, and of GSC + GDRX + CDRX are presented below.

These results were deemed reasonably accurate, and the microstructure model was then applied to the case of indirect extrusion. Various points on the extruded billet were tracked during the FEM simulation, and the state variable histories of temperature, strain, and strain rate were exported. These histories were provided to the microstructure evolution model, which then “kept track” of the features listed above.

Employing the same analysis, the ratio of grain boundaries and subgrain boundaries was determined, and ultimately the final grain size computed. This is compared to the final grain size measured in the DMZ of the indirectly extruded billets in Figures 10 and 11. The comparison is good, although a better comparison may be made if more microstructures are observed at a wider range of strains, for both the torsion tests and the indirect extrusion experiments.

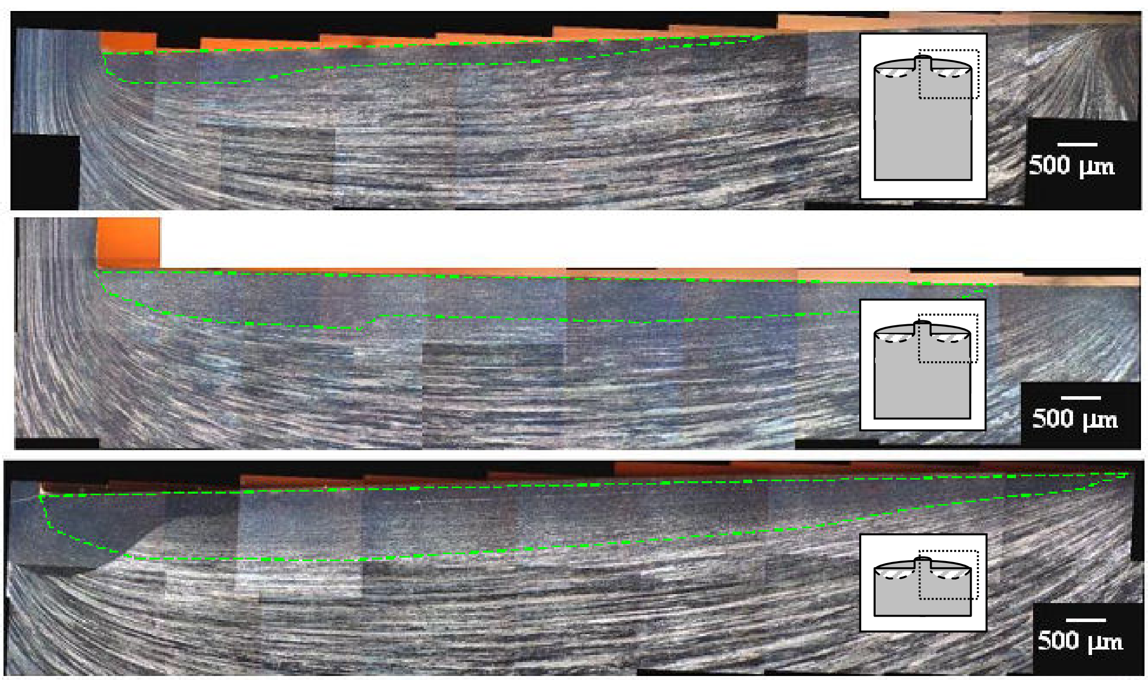

Figure 10. Photomontages of partially extruded indirect billets. Equiaxed, fine grains within the "DMZ" of these billets are highlighted by dashed green lines. Top - billet extruded 25% of its length Middle - billet extruded 50% of its length. Bottom - billet extruded 75% of its length. Inset figures denote the region of each split, etched, partially extruded billet which was observed to characterize DMZ.

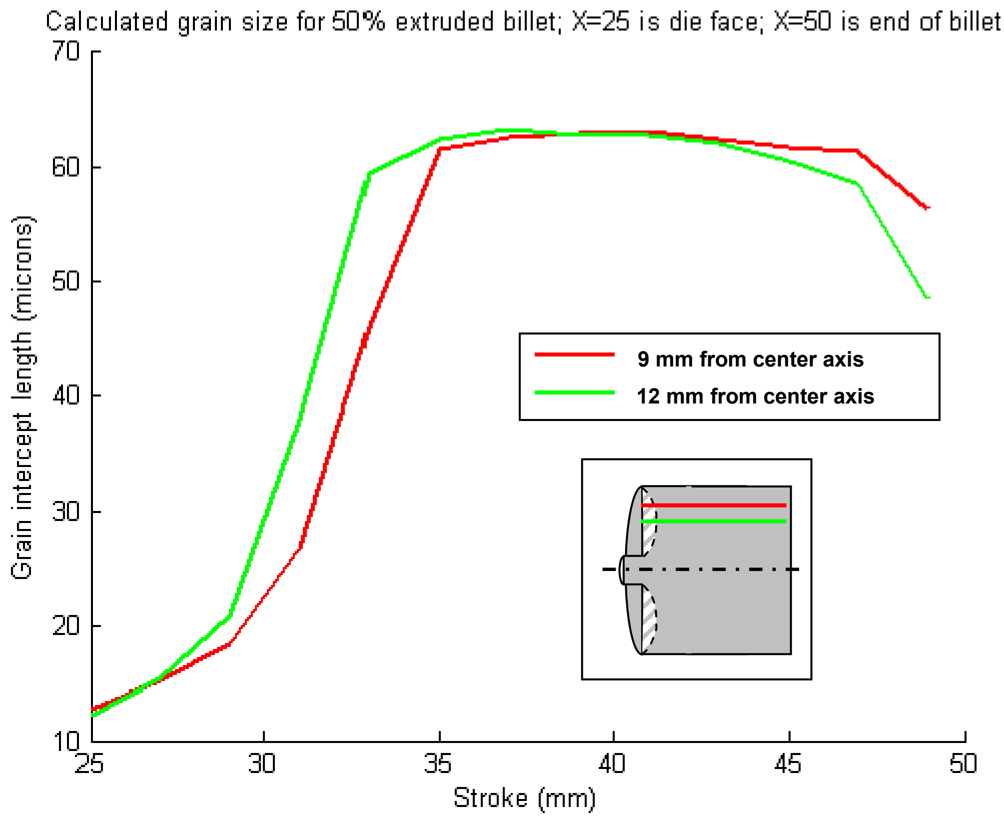

Figure 11. Microstructure modeling predictions of grain size (represented by average intercept length) within a partially indirectly extruded billet, as a function of distance from the die face. Inset figure indicates where grain intercept lengths (proportional to grain size) were predicted with respect to the position within the partially extruded billet.

DISCUSSION

The microstructure model predicts a transition from the large, initial grains to fine, recrystallized-like grains in the region close to the die face. This is indeed similar to the phenomenon of fine, equiaxed grains in the DMZ-region of the experimentally observed indirectly extruded billets. However, the transition point from large to fine grains comes much further back in the billet (further from the die face) in the model prediction than observed experimentally. Additionally, the gradient in grain size is less steep in the model than observed experimentally. This may be due either to underpredictions of strain in the FEM model, imperfect material flow modeling in the FEM model, overpredictions in the subgrain size calculations, and imperfect modeling of the TO distortion in the Matlab™ program. This last option is a significant probability, given that the TO model was modeled only in one particular orientation. That is to say, the computation of the distortion of the TO was performed with the deformation axis (whether it be tension or compression) in just one axis of the TO. In reality, however, assuming that grains can be accurately modeled by a TO, they would be oriented in many random directions with respect to the deformation axis. Since the kinetics of GDRX occur as the subgrains break out of the “thinned” portion of the parent grain, and since this occurs first at sharp-faceted edges, if the deformation axis were aligned along one of these sharp facets, rather than aligned with a flat face, the kinetics would be slightly different. To properly capture the average kinetics of GDRX for a large collection of grains, the orientation of the TO should be varied with respect to the deformation axis. DePari performed an analysis of the difference in GDRX kinetics based upon deformation along two different axes[14], and indeed showed that they vary depending on the axis on the TO along which deformation occurs.

Furthermore, a finer sampling of the ratio of subgrains to grains as a function of strain, both in the validating torsion test and in the indirect extrusion test, is necessary to better understand the correlation between the model and experiments.

Applications to the Peripheral Coarse Grain Defect

Lastly, modeling of the recrystallization and grain growth kinetics of these fine, equiaxed grains which flow into the extrudate, and which subsequently may become the PCG, must be validated. Modeling efforts using both voronoi cells and cellular automata to predict this phenomenon have been performed[3], but are beyond the scope of this paper and will be presented elsewhere. The issue to note is that these fine, equiaxed grains formed in the DMZ-region of the indirectly extruded billets, slip into the extrudate, and apparently grow into the large, coarse grains of the PCG. The fact that they are probably refined due to GDRX, rather than a more conventional recrystallization phenomenon (either metadynamic or static) provides a reason why they may bloom into such large, coarse grains later. MRX and SRX are both recrystallization phenomena which follow the mechanism of "nucleation and growth" – that is to say, sufficient dislocation density and disorder accumulate during deformation that a new, recrystallized, relatively dislocation-free grain nucleates, and a driving force exists to consume the higher energy region around it. This would leave a fine, equiaxed microstructure with relatively low dislocation density – no driving force would exist to compel it to grow again later in the extrudate, as the PCG appears to do. However, grains refined via GDRX are qualitatively different from grains refined via other recrystallization phenomena, as they still represent grains of very high dislocation density. This means that they are prime for subsequent energy reduction via a post-deformation recrystallization mechanism such as metadynamic or static recrystallization.

Note - while the results presented here apply to indirect extrusion, the theory and associated mathematical relationships could be applied to direct extrusion as well.

CONCLUSION

In sum, it appears highly plausible that grains flowing along the exterior of an indirectly extruded billet experience a very large amount of strain, as depicted by FEM analyses validated with visioplastic studies. This strain is sufficient to create grain break-up via the GDRX phenomenon, resulting in fine, equiaxed, recrystallized-like grains with high dislocation density at the die face, which then slip into the periphery of the extrudate. Given that these grains were refined mechanically, rather than via a nucleation and growth mechanism, they still possess a very high dislocation density – the driving force necessary to induce post-deformation recrystallization, such as metadynamic or static recrystallization. This may be the phenomenon occurring a few seconds after the grains slip out of the orifice, while still at high temperature and at high energy. The distinct PCG layer exists because grains flowing from the center of the billet (as opposed to those slipping in from the DMZ-region) experience less deformation, and/or have lower grain boundary density, and thus a lower density of nucleation sites for subsequent recrystallization.

Future work should include further analyses of the extrudate/billet near the die face, such as microhardness testing across the extrudate just past the die orifice, to determine if a distinct difference in dislocation density/properties exists; EBSD analyses to determine the density of grains/subgrains (possible nucleation sites); and/or TEM analysis to quantify the dislocation density difference in these regions. Such studies would further help to verify if this is indeed the mechanism responsible for the PCG formation, and if so, spark a discussion as to how to tailor the extrusion process to reduce this defect.

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