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Entitled

Extrusion Machining: A Thermo-Mechanical Process for Producing Strips of Alloys Having Limited Workability

For the degree of Doctor of Philosophy

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EXTRUSION MACHINING: A THERMO-MECHANICAL PROCESS FOR PRODUCING
STRIPS OF ALLOYS HAVING LIMITED WORKABILITY

A Dissertation

Submitted to the Faculty

of

Purdue University

by

Mert Efe

In Partial Fulfillment of the

Requirements for the Degree

of

Doctor of Philosophy

August 2012

Purdue University

West Lafayette, Indiana

ACKNOWLEDGEMENTS

I would like to begin with the funding sources which in part by made this work possible. Funding sources are listed below in chronological order that they have received:

1. National Science Foundation (NSF), Civil, Mechanical and Manufacturing Innovation (CMMI) grants No: 0800481 and 1100712, Program Manager: Dr. Mary Toney.
2. U.S. Department of Energy, Pacific Northwest National Laboratory grant: DE-AC06-76RL01830, Program Managers: Mark Smith and Dr. Aashish Rohatgi.
3. Purdue University, Bilsland Dissertation Fellowship.
4. U.S. Army Research Office grant: W911NF-12-1-0012, Program Manager: Dr. Suveen N. Mathaudhu.

I would like to personally thank to our program managers Dr. Aashish Rohatgi and Dr. Suveen N. Mathaudhu for their support, discussions and professional networking.

Many thanks to my advisors Prof. Kevin Trumble and Prof. Srinivasan Chandrasekar for their guidance, encouragement and suggestions regarding my research. I have spent wonderful 6 years at Purdue due to their relaxed yet supportive attitude. I think, I have become an independent researcher due to the free and unbounded research environment provided by my advisors. I would like to follow their style of mentoring if I pursue a career in academia. I would also like to thank my committee members Prof. Keith Bowman and Prof. Jean Paul Allain for their time and suggestions on this thesis.

This work was a continuation of the research started by Dr. Wilfredo Moscoso. He had introduced the concept of extrusion machining (completely independent of the earlier works of Dr. Leonardo De Chiffre) and laid out the fundamentals of the process. I have very much enjoyed working with him for over 4 years and applying extrusion machining to alloys with limited workability. I have learned so much from him on being a systematic researcher. Moreover, he and his family (Mabel, Beatriz and Marcel) have become my second family here in West Lafayette. I am truly indebted to the friendship, family comfort and warmth provided by them.

My other fellow research companion, Dinakar Sagapuram, has helped me in many aspects of the research. Many of the micrographs, EBSD and x-ray data presented in this thesis were produced by him. He has been very courteous with providing and sharing the data. I have also benefited his most insightful suggestions for improving the discussions presented in Chapter 6. He and his girlfriend Neha, together with Dr. Moscoso and his family, have also provided me a great friendship.

I would like to thank to my other friends in Center for Materials Processing and Technology for their assistance. Dr. Yang Guo provided the PIV data and contributed to the process characterization. Dr. Chris Saldana (currently at Penn State) and Sergey Suslov helped with the TEM sample preparation and microscopy.

Dr. Patricia Metcalf, David Meyer and Materials Engineering staff provided invaluable assistance during my research. Patti's help and advice with the furnaces, cutting procedures and almost everything has never gone unappreciated. Special thanks to Wayne Ewbank for helping me with the everyday machining activities.

Thanks to all my other colleagues in Materials Engineering and all my friends here in West Lafayette and my family for their support and company during my research. Last, but not least, I thank Idil Aksoz for being with me for the last 4 months of my

graduate work. She has unconditionally supported me during my thesis writing process, endured all my nags and even helped me writing the references section.

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ABSTRACT

Efe, Mert. Ph.D., Purdue University, August 2012. Extrusion Machining: A Thermo-Mechanical Process for Producing Strips of Alloys Having Limited Workability. Major Professor: Kevin P. Trumble.

Deformation processing of alloys with limited workability is one of the principal technological challenges for the metal manufacturing industry. Traditional processes (rolling, extrusion) are usually performed at high temperatures ($0.5 T_m$) for increased plasticity and crack-free products, which makes them costly, energy intensive and raw material inefficient. In this thesis, extrusion machining (EM) a shear-type deformation process is applied to alloys with limited workability for sheet and foil production from bulk forms in a single step. EM is shown to have a narrowly confined deformation zone with much wider control of deformation than in conventional deformation processing (e.g., severe plastic deformation). Controllable deformation parameters include strain, strain rate, hydrostatic pressure, temperature and deformation path. These attributes are highlighted in deformation processing of two classes of limited workability alloys: 1) Alloys with intrinsic susceptibility to cracking (and segmentation) due to flow localization and shear band formation and 2) Alloys with extrinsic susceptibility to cracking due to macroscopic defects such as porosity, casting defects and weak second phases. Magnesium AZ31 alloy is selected as a model material for the first class and sintered copper, cast brass alloy 360 and cast Mg AZ31 are selected for the second class. Noteworthy features of the process are: *suppression of segmentation* by combinations of high hydrostatic pressures ($p/2k > 1.2$) and deformation temperatures ($T > 0.5 T_m$) realized by in-situ plastic heating, realization of a range of strains $0.9 < \bar{\epsilon} < 1.2$ and deformation rates $10^2 < \dot{\epsilon} < 10^5/s$, *engineering of*

microstructures ranging from conventional to ultrafine grained, *shear type deformation textures* (non-basal in the case of Mg AZ31) and creation of sheet from the bulk *in a single step* of deformation *without pre-heating*. The thermo-mechanical conditions and microstructural processes (e.g. continuous dynamic recrystallization for Mg AZ31) that are responsible for continuous and sound sheet products are established. Guidelines for scalability and cost-effectiveness of EM for commercial production are analyzed and discussed.

CHAPTER 1. INTRODUCTION

Deformation processing of alloys having limited workability has been a challenge for the metal manufacturing industry for quite a while. These alloys demand to be processed at high temperatures (typically at or above $0.5 T_m$) and high pressures, which are critical for increased plasticity and crack-free products [1-4]. Even at these high temperatures, multiple deformation steps and slow production speeds are also required. As the thickness decreases during the cold rolling of sheets or foils, the number of deformation steps and the number of intermediate annealing steps is increased to achieve the final cross-section. In addition to these steps, slow production speeds sum up to relatively expensive and challenging production.

Traditionally, rolling and extrusion have been employed as the conventional techniques for the production of flat and long products. In order to achieve the desired temperatures, the workpiece and the tools are preheated and heat loss becomes a significant problem due to the slow production speeds and large tool sizes which act as large heat sinks. Additionally, these processes usually apply incremental deformation during each pass and multiple steps are inevitable for the desired shape change. Often the workpiece needs to be heated between each step to keep the temperatures at the desired level. Surface contamination and oxidation become additional issues when the workpiece of choice is reactive at high temperatures. The contaminated surface and the oxide scale need to be scalped or pickled after each processing step, which significantly decreases the yield of the conventional processes. Consequently, these processes become inefficient and expensive when deforming alloys having limited workability into strips (sheets with limited width) or foils.

Therefore, a low-cost manufacturing process is needed for the alloys having limited workability in order to bring down the costs of the final forms with the smaller cross-section. An extrusion machining (EM) technique has been described before to produce sheets, strips and wires of metal alloys in a single step [5, 6]. A variation of this process –large strain extrusion machining (EM) - has been also shown to produce strip, sheet and wire forms of OFHC copper and 6061 aluminum alloy, with ultrafine and even nanocrystalline microstructure [7-9]. Compared to rolling or conventional extrusion, extrusion machining is a natural high strain rate process (strain rates typically above $10^2/s$) due its narrow deformation zone. The requirement for workpiece and tool pre-heating is minimal due this narrow zone as well. The deformation heat stays in the deformation zone and in the workpiece at these high strain rates and heat loss becomes less of a problem as the speed of the process can easily be increased. Moreover, hydrostatic pressure can be easily applied and controlled during this process by the proper design of deformation zone geometry. This geometry is controlled by tool design and related process parameters. Consequently, this local temperature rise and the application of hydrostatic pressure is an efficient and fast process to help plasticity and required shape change, which can even be accomplished in a single step for alloys directly from the as-cast microstructures. In this thesis, extrusion machining is proposed as a low-cost method to produce foil, strip and sheet forms of alloys with limited workability.

The extrusion machining process is applied to two major classes of alloys with limited workability: 1) Alloys with intrinsic susceptibility to cracking (and segmentation) due to flow localization and shear band formation and 2) Alloys with extrinsic susceptibility to cracking due to macroscopic defects such as porosity, casting defects and weak second phases. Mg AZ31B alloy is selected as a model material for the first class due to its susceptibility to shear band formation at high temperatures and strain rates. Sintered OFHC copper , leaded cast brass alloy 360 are examined in the second class due their extrinsic susceptibility to cracking. The

macroscopic defects are the pores from sintering in the case of copper and weak second phases from casting in the case of 360 brass. Cast Mg AZ31 alloy is also studied under the second class, in order to directly compare the processing behavior of wrought and cast materials.

The Mg AZ31B will be the main focus of the thesis since it is a perfect model material to study the change in plasticity with the selected process parameters. Magnesium shows a definite transition in deformation mechanism and remarkable increase in plasticity around 200 °C at low strain rates [10]. This transition can be easily studied by analyzing microstructures; hence the process and its parameters that result in continuous deformation and crack-free forms will be studied with the microstructures and mechanical models.

The deformation zone in the extrusion machining process was analyzed before by various researchers [5, 9]. Strains and hydrostatic pressures in the deformation zone of the process were modeled and will be reviewed here. The temperature rise in the deformation zone is modeled with the help of measured cutting forces and energies. Therefore, the extrusion machining can be treated as a thermo-mechanical process where strain, strain rate, pressure and temperature can be controlled, confined and accurately modeled.

In order to draw the complete picture of the process, shear localization models are also applied to the Mg AZ31B alloy. Since extrusion machining is a high strain rate process, adiabatic shear banding due to the thermal softening is the main reason for segmentation and cracking. However, it is found that shear localization is a problem up to a critical processing speed or strain rate. Once this critical speed is achieved, local temperature rise is sufficient enough to change the deformation mechanism, thereby resulting in crack- and segmentation-free foils, strips and sheets. Shear localization models are revised based on these findings.

Consequently, the aim of this thesis can be summarized as the mapping of the thermo-mechanical conditions and microstructural processes that are responsible for continuous and sound sheet/strip/foil making of limited workability alloys by extrusion machining.

1.1 Extrusion Machining vs Conventional Machining (Peeling)

Figure 1.1 shows a schematic of the plane strain EM used in the present study. A sharp, wedge-shaped tool removes a preset depth (initial chip thickness, t_o) and width (t_w) of material in the form of a chip-extrudate by shear at rate V_o (also referred as V for simplicity) (see Supplementary Video 1 on extrusion machining of pure copper). The shear is confined to a deformation zone idealized as a shear plane. In contrast to conventional machining, the geometry of the chip in EM is fixed at formation by using a constraining edge across from the tool. The constraint is designed so that the cutting and extrusion occur simultaneously, converting the machining-chip formation into a controlled deformation process. The other important parameters of the process are: t_c - the final chip (extrudate) thickness, λ - the chip thickness ratio (t_c/t_o), α - the tool rake angle and Δ - the thickness of the deformation (shear) zone.

In contrast to conventional machining, the extrudate thickness (t_c) is set *a priori* before the machining is initiated. Also, t_c can be set smaller than t_o , enabling large plastic strains and high hydrostatic pressure levels to be imposed concurrently in the deformation zone. The inset to Fig. 1.1 shows an image of EM, obtained by averaging 200 consecutive frames from a high-speed image sequence, which highlights details of the flow such as pathlines and division of the flow into chip and workpiece. Since the flow is steady -state, the pathlines are also streamlines whose tangents are the velocity vector directions. The chip product in EM is a sheet with geometrical attributes akin to those produced by plane strain rolling or extrusion. Furthermore, the sheet is produced directly from the bulk material in a single stage of deformation.

Extrusion machining or simple machining have never been tried before as methods for producing strips/foils of alloys with limited workability. The physics and mechanics of continuous strip formation need to be fully understood. Segmentation due to adiabatic shear banding were well understood as the reasons for discontinuous strips (chips) in flow softening alloys (Class 1), but the previous works were focused on the segmentation phenomenon in order to achieve broken chips from these alloys. Continuity of the chips was never a problem, rather the interest was on the surface finish of the workpiece and the economics of the process [12, 13]. Therefore, segmentations were never controlled before in order to get continuous products. The segmentation phenomenon has been studied in the equal channel angular pressing (ECAP) literature for the sake of getting continuous products from the process [14, 15], but the ECAP was never identical to the problem of extrusion machining.

A detailed study of chip segmentation mechanics, in the context of materials fitting category (1) above, is proposed in order to correlate appropriate deformation parameters with the physics governing the phenomena. The importance of such a study lies in that uncontrollable chip segmentation leads to defective (cracked) sheets/foils and non-homogenous microstructure & properties.

The study will include experimental validation of existing analytical models and development of modifications where experiments contradict theory, as well as incorporation of the effect of applied hydrostatic pressure in EM. The effect of applied pressure has also never been studied in the context of conventional machining. The models may be used to predict EM parameters leading to defect-free sheets/foils for any material in category (1) above. The existing analytical models for segmentation (adiabatic shear banding) assume constant material properties and deformation mechanisms over the range of the processing parameters. Although this might have been true for the materials previously studied in the literature, magnesium showed an exceptional behavior. The validity of the models

for the materials that show different flow softening behavior stands out as a question in the literature and this issue will be addressed through the magnesium study. The exceptional behavior of magnesium will be explained with the change in deformation mechanisms.

CHAPTER 2. EXTRUSION MACHINING AS A THERMO-MECHANICAL PROCESS

2.1 Historical Development of the Peeling Process and the Need for Extrusion Machining

In conventional (free) machining (FM) or peeling, where there is no constraining tool, λ is developed naturally depending on the feed, rake angle and the cutting speed. Then in the case of the extrusion machining, λ is controlled with the application of the active constraining tool. The final thickness of the chip can be decreased below natural λ and the exit geometry is controlled. However, for a given set of machining conditions the maximum λ and the final chip thickness attainable is still the same as the natural λ .

A simple peeling operation, which resembles the extrusion machining with an inactive constraining tool, can be traced back to late 1800s [16] (Figure 2.1). The aim of the inactive constraining tool seems like to control bending or curling of the strip and to provide guidance to the strips. Historically, this process appears to be first developed in 1862 in the form of peeling without any constraining tool by Kinder [17] in order to produce tin and lead strips. Later the process came into attention of many inventors, either in the forms of peeling or peeling with inactive constraining tool. Even commercialization attempts have been carried out by large corporations such as General Motors, British Steel and Gillette [18-20]. They all used the peeling process with applied chip tension. The tension to the strip is provided with a coiler which operates faster than the machining process. The applied tension is found to decrease natural λ (final chip thickness) and the cutting forces. Tool wear is reduced via the decrease in the cutting forces. Roughness of the free surface is also decreased with the tension [21]. However, application of tension does not

supply any hydrostatic pressure and does not have any advantage in deforming alloys with limited workability.

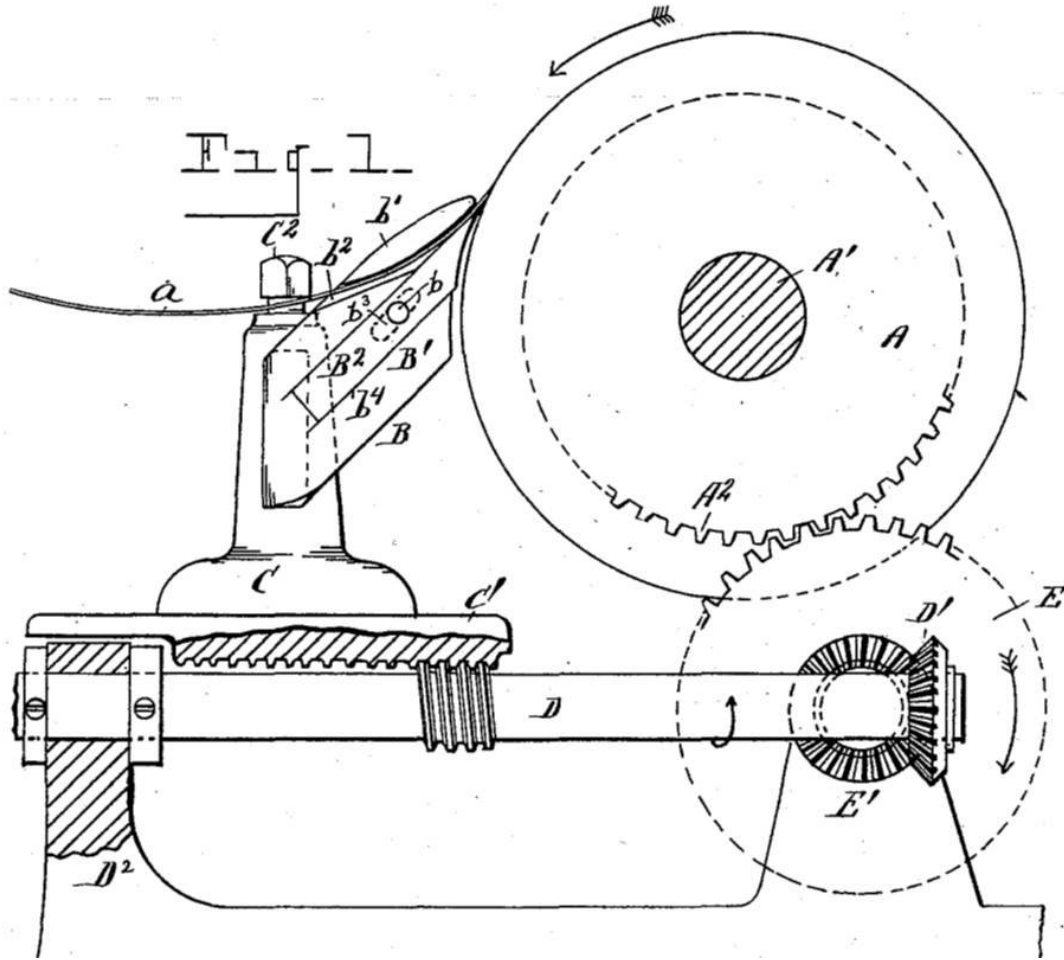


Figure 2-1 A historical peeling process for metallic strip/foil production. The constraining tool (b') is thought to be inactive in this case. It seems like the aim of this tool is to provide guiding to the strip rather than extruding (after [16]).

These commercialization attempts of the peeling seem to be failed, unfortunately, since there are no known current applications of the process. Even with the application of chip tension, the roughness of the free surface was a problem in peeling [20]. Strips had to be uncoiled and cold-rolled in order to decrease the surface roughness and to produce an acceptable product. Then they have to be

recoiled again, which was one of the extra processing steps. Extrusion machining provides a solution to this problem by peeling and extruding the strip simultaneously, resulting in smooth finishes on both surfaces of the strip.

A considerable amount of work had been spent on tooling and machine design for the peeling process, but materials selected for the process were those demanding high machining forces such as steels, stainless steel, copper and brass [20]. The need for expensive tooling, machinery and capacity requirements might hinder the commercialization. These materials are also easy to work with the conventional rolling. Especially, copper and brass offer significant amount of plasticity even at room temperature and at ordinary strain rates. The peeling may not have offered enough cost advantage, even at the smallest cross-sections (thicknesses).

If the focus had been on the rather soft metals and alloys such as cast aluminum and magnesium and pure titanium, a niche market for the peeling could have been developed. Tool wear and capacity requirements would have been lower for these materials. The advantage of the narrow deformation zone of the peeling process would have been fulfilled and the local temperature rise would be helpful to inhibit crack formation and segmentation. Peeling becomes ideal where the material to be worked shows sufficient plasticity only at high temperatures and pressures. Even then, a need for extrusion machining arises since only it can provide the sufficient hydrostatic pressure that is essential to form these limited workability materials.

In order to make the extrusion machining economically feasible, the material should also not demand high cutting forces; hence expensive tooling, machinery and capacity requirements. As a result, the process becomes feasible below some certain thicknesses of sheets, strips and even foils where conventional processes require multiple steps of deformation, annealing and pickling, expensive production and smaller yields.

The starting material (feedstock) cost is the other most important aspect of the process economics. Starting material constitutes more than 50% of the production cost. Previous peeling processes used expensive wrought materials (rolled or forged) as the feedstock and some researchers experimented with cheaper cast feedstock [20]. It was concluded that the peeling process did not offer significant cost savings when wrought forms are used as the starting material due to the fact that wrought forms are already processes with hot forming (hot rolling, forging or extrusion) techniques and the peeling process only eliminates cold forming steps. In order to make peeling or extrusion-machining cost effective, cast materials must be used. Moreover, the type of the casting operation affects raw material costs. For this reason, the cheapest casting technique needs to be employed for the feedstock production. Only in that ways, both the hot rolling and cold rolling steps can be truly eliminated and material and cost saving features of the peeling or extrusion machining processes can be obtained.

Using cast materials as a starting material is, of course, more challenging than wrought materials. A typical cast material may contain large porosity, weak second phases and inhomogeneity in grain size. Therefore a cast feed stock will fit into the category of alloys of limited workability due to the extrinsic susceptibility to cracking/segmentation. Extrusion machining, again, provides the advantage of hydrostatic pressure for suppression of segmentation as opposed to peeling.

Extrusion machining with an active constraining tool appeared to be first developed by De Chiffre [5]. Continuous strips of brass were successfully produced. Later, Moscoso took an extra step and investigated the process more for especially high strains [7]. Continuous forms of OFHC Cu and Al 6061 alloy were produced at shear strains up to ~ 10 , resulting in ultrafine grained structures by taking advantage of the high-strain nature of the process.

2.2 Mechanics of the Process

2.2.1 Deformation Strain and Strain Rate

Figure 2.2 shows a typical deformation field in EM obtained using Particle Image Velocimetry (PIV), an image correlation technique. PIV involves recording of sequences of images of the deformation zone at high framing speeds, see inset to Fig. 1.1, and analysis of the images to track motions of particle ensembles (markers) and material flow. Displacement, velocity, and strain rate and strain fields associated with the deformation are estimated by the image analysis [22, 23]. The effective strain rate ($d\bar{\epsilon}/dt$) field is shown in Fig. 2.2a, while strain rate and strain profiles along a particle trajectory $abcd$ traversing the deformation zone and entering the sheet are shown in Figs 2.2 b&c, respectively. The strain rate, a measure of the strain increment imposed during an incremental advance of the tool, provides a basis for characterizing the deformation. The region of intense strain rate is the deformation zone (bc); it is seen to be quite confined and uniform with average thickness, Δ , of $\sim 55 \mu\text{m}$ (Fig. 2.2). The EM thus imposes intense shear deformation (i.e., severe plastic deformation - SPD) in the sheet in a rather narrow zone extending from the cutting to the constraining edge. This confinement of the deformation results in a localized temperature field [24]. If the deformation zone is idealized as a shear plane, then the shear strain, γ , in the sheet is given by the upper-bound models that are applied to extrusion machining by De Chiffre [5]:

$$\lambda = \frac{t_c}{t_0} \quad (1)$$

$$\gamma = \frac{\lambda}{\cos\alpha} + \frac{1}{\lambda\cos\alpha} - 2\tan\alpha. \quad (2)$$

λ and α are previously defined as the chip thickness ratio and cutting tool rake angle, respectively. The effective strain is $\bar{\epsilon} = (1/\sqrt{3})\gamma$ (referred as simply ϵ from now on). This model can be applied to free machining, however λ in that case is not a controlled variable but it is a process outcome.

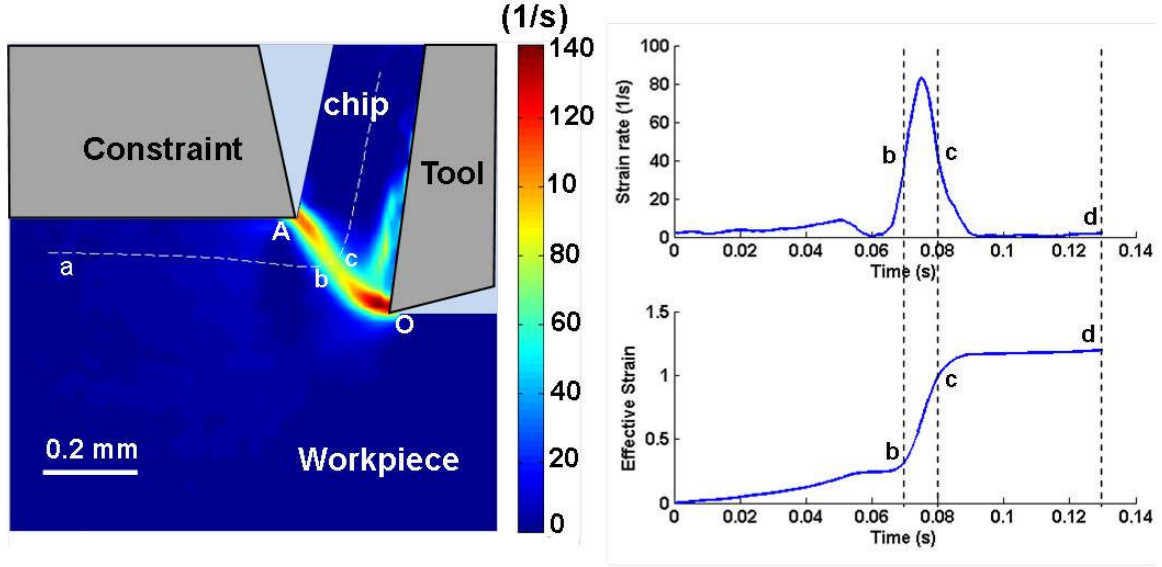


Figure 2-2 PIV analysis of deformation showing a) strain rate field b) effective strain profile rate along abcd and c) effective strain along abcd. Lead, $\alpha = +10^\circ$, $t_o = 0.2$ mm, $V_0 = 10$ mm/s, $\lambda = 1$. The thickness (Δ) of the deformation zone (bc) is ~ 55 μ m (after [11], used with permission of Elsevier).

Based on the PIV, a representative value for $d\varepsilon/dt$ in the deformation zone is obtained by spatially averaging the rate data in this zone. This gives $d\varepsilon/dt \approx 70/s$ for the condition of Fig. 2.2. PIV measurements have also shown $d\varepsilon/dt \approx \varepsilon V_o/\Delta$; thus the strain rate in EM can be varied from 20 to $10^5/s$ via V_o (~ 1 mm/s to 5 m/s).

The strain is obtained from the PIV strain rate data by integration along particle trajectories (e.g., abcd in Fig. 2.2) [22, 23]. The effective strain in the sheet thus estimated is ~ 1.2 for this deformation condition ($\lambda = 1$, $\alpha = +10^\circ$) as against a value of $\varepsilon = 0.97$ from the shear plane model (Eq. 2). Close correspondence between PIV measurements and shear plane model estimates of strain has been noted for λ down to 0.7. For $\lambda < 0.7$, the PIV measurements, while still following the trend of Eq. 2, give much larger values for the strain; for example, at $\lambda = 0.4$, the PIV strain is about 80% larger than that given by Eq. 2. In the present study, where the EM conditions use $\lambda \geq 0.7$, the shear plane model for γ values (Eq. 2) is used.

Figure 2.3 shows the variation of γ with λ for select α (Eq. 2). For a given α , γ is a minimum at $\lambda = 1$; in the neighborhood of $\alpha = 0$, this minimum value of γ is ~ 2 . Reducing λ below 1 or increasing it above 1 increases γ . The power capacity and machine system stiffness, together with material characteristics, determine the minimum λ (lower limit), since forces increase with decreasing λ for $\lambda < 1$. Chip formation characteristics of conventional machining determine the upper limit or maximum λ . In ductile metals such as copper, λ can be set as large as 10, giving $\varepsilon \sim 6$ for $\alpha = 0$ [22].

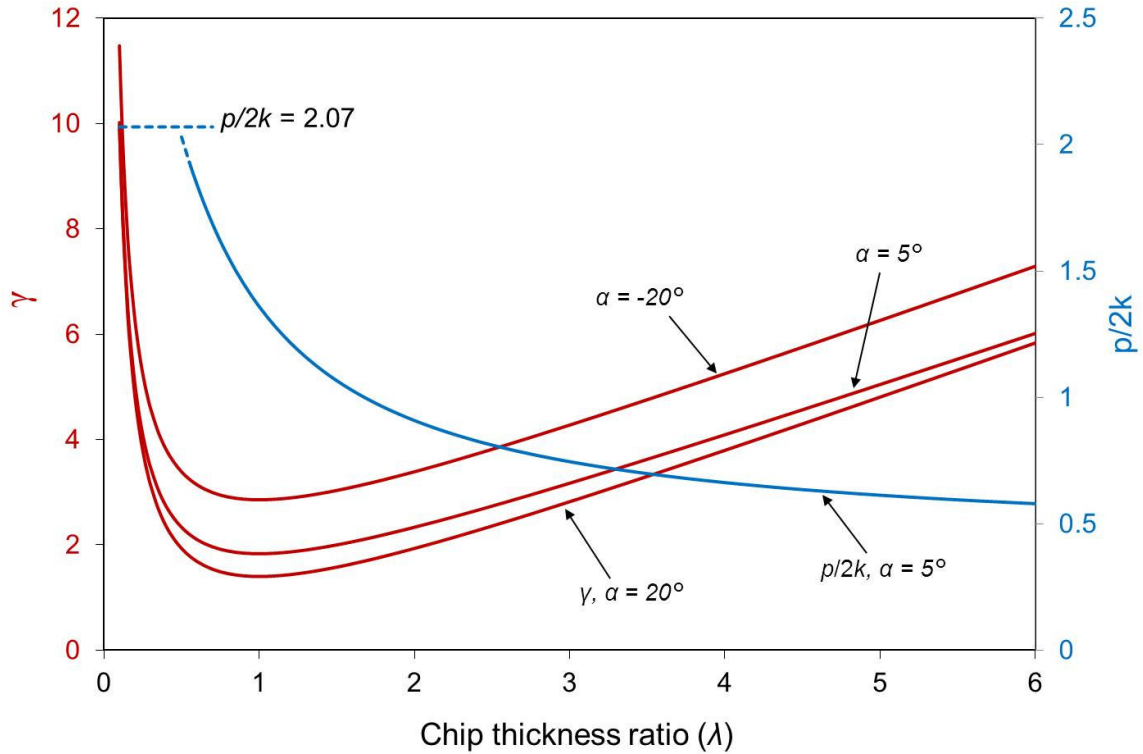


Figure 2-3 Variation of shear strain (γ) and normalized hydrostatic pressure ($p/2k$) with chip thickness ratio (λ) and rake angle (α). While γ shows a minimum at $\lambda = 1$, the pressure increases with decreasing λ up to a maximum of $p/2k \sim 2.07$ (plane strain punch indentation) (after [11], used with permission of Elsevier).

2.2.2 Slip-line Field for Extrusion Machining and Hydrostatic Pressure

Moscoso has developed a slip-line field model for extrusion machining [9]. Slip-line field model overcomes some disadvantages of the upper-bound model and allows hydrostatic pressure to be derived. Below is the application of slip-line fields to the case without a built-up edge on the cutting tool (Figure 2.4). This is a more realistic case where there is a secondary shear zone right at the tip of the cutting tool due to the friction between the workpiece and the cutting tool.

The model [9] assumes plane strain deformation, rigid perfectly-plastic material where there is no strain hardening, isotropic properties, steady-state plastic deformation and no friction between the workpiece and the constraining tool. The

plane strain assumption is valid when deforming sheets or strips of sufficient width and thickness ($w/t \gtrsim 5$), and the perfectly plastic case is also sometimes valid when the deformation zone temperature is high enough to satisfy dynamic recovery and recrystallization. Isotropic properties (such as yield strength - σ_{ys}) is, of course, not a realistic assumption for wrought metals and alloys. However, slip-line fields still model the process more accurately than the upper-bound model and will be used further in this thesis for pressure calculations, since it is possible to obtain hydrostatic pressure from slip line fields.

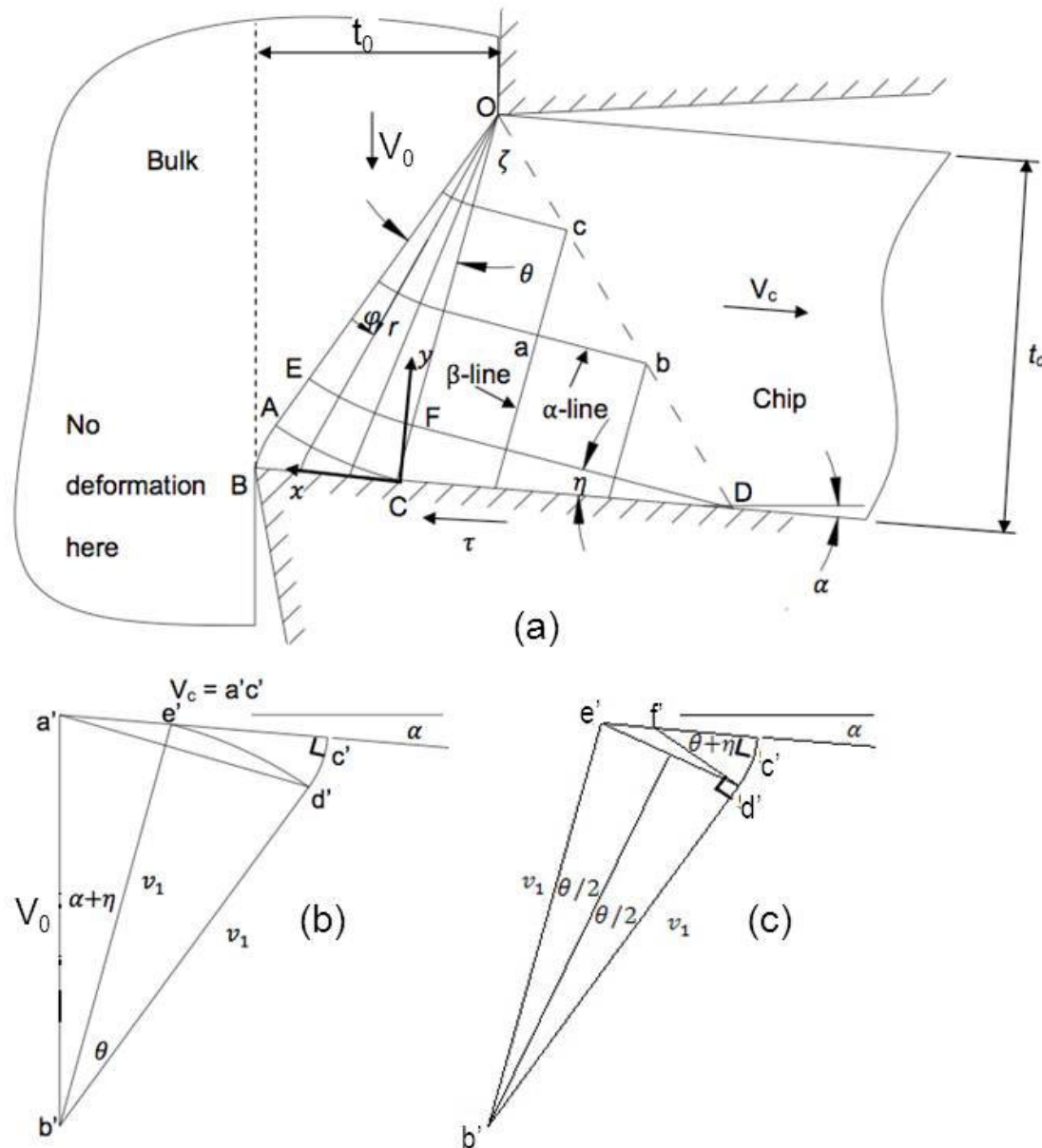


Figure 2-4 a) SLF for plane-strain EM., b) velocity hodograph and c) auxiliary construction to obtain θ . The work material is assumed rigid-perfectly plastic (after [9, 11], used with permission of Elsevier).

From the Figure 2.4, the maximum hydrostatic pressure and the total strain can be calculated once the deformation zone angle (fan angle) θ is known [9]. In the case of maximum friction between the tool and the workpiece, where τ_F is the friction

stress on the tool face and equals the shear strength (k) of the material (i.e., $\tau_F = k$), the solution for θ reduces to:

$$\theta = 2\tan^{-1}\left(\frac{1}{2}\left[\frac{1}{\lambda\cos\alpha} - \tan\alpha\right]\right). \quad (3)$$

Then the total strain at the exit of the deformation zone is:

$$\gamma = \gamma_1 + \gamma_2 = \frac{1}{\sin\theta} + \ln\left\{\sec\left(\frac{\theta}{2}\right) + \tan\left(\frac{\theta}{2}\right)\right\} \quad (4)$$

where γ_1 and γ_2 are the primary and secondary shear zone contributions, respectively. The two models agree well on the strain for $\lambda > 1$, however there is a slight decrease in the strain predicted by slip-line fields for $\lambda < 1$ compared to the strain by upper-bound model. Also the minimums of strains with λ are slightly different. The SLF model predicts the strain more accurately than the upper-bound model [9].

The maximum hydrostatic pressure, p , is achieved at the entry of the deformation zone and given as [9]:

$$\frac{p}{2k} = \frac{1}{2} + \theta \quad (5)$$

For zero rake angle tool ($\alpha = 0$), the above equation reduces to:

$$\frac{p}{2k} = \frac{1}{2} + 2\tan^{-1}\left(\frac{1}{2\lambda}\right) \quad (6)$$

where hydrostatic pressure is only a function of λ . The pressure is normalized with the uniaxial yield strength ($2k$) of the material.

The hydrostatic pressure predicted by slip-line fields is drawn as a function of the λ for the zero rake angle tool (Figure 2.3). Figure 2.3 shows the variation of $p/2k$ with λ for $\alpha = 5^\circ$, the rake angle commonly used in the experiments. The pressure decreases with increasing λ (Fig. 2.3). It can be set potentially as small as $\sim 2k$, the shear strength. The maximum value for $p/2k$ is ~ 2.07 , the value corresponding to

plane strain punch indentation. Process conditions of $\lambda < 1$ provide unique combinations of large ε and high p - conditions well-suited for effecting microstructure refinement and deformation processing of alloys of poor workability.

Although it is possible to increase both strain and hydrostatic pressure at the same time by decreasing λ , experiments are showing there is a limit of minimum λ obtainable for each material governed by the onset of side flow. As a general rule of thumb, the minimum λ obtainable increases as the natural λ for that material increases. In other words, if the workpiece shows significant ductility during any forming operation, particularly in free machining, the application of the hydrostatic pressure by the constraining tool becomes unrealistic and infeasible. In the case of ductile metals (for example: OFHC copper or wrought unalloyed aluminum), there will be side flow and material will not flow through the opening provided by the tool, especially for $\lambda \ll 1$.

2.3 Shear Energy in Machining and Temperature Rise Modeling

Temperature rise in any of the deformation processes can be given through the amount of deformation work converted to heat under adiabatic conditions:

$$dT = \frac{\beta \sigma d\varepsilon}{\rho C_p} \quad (7)$$

where β is the fraction of plastic work converted to heat (typically 0.95), σ is the flow stress, ε is the plastic strain, ρ is the density and C_p is the specific heat of the given material [12]. The total temperature rise in the deformation zone due to adiabatic heating is:

$$\Delta T = \frac{\beta}{\rho C_p} \int_0^{\varepsilon_f} \sigma d\varepsilon. \quad (8)$$

The same type of a relation is applied to the case of free machining in order to predict the temperature rise;

$$\Delta T = \frac{\beta_1 U_s}{\rho C_p} \quad (9)$$

where β_1 is now a coefficient that includes both the fraction of the deformation energy transferred into heat and the fraction of the heat generated that stays in the deformation zone as opposed to the heat lost to the workpiece. In other words, β_1 is the adiabaticity factor and as adiabaticity increases the heat generated by the deformation stays in the deformation zone rather than diffusing into workpiece (Figure 2.6). In the case of machining, the specific shear energy, U_s (J/m³), is a function of specific shear energy rate, U_{sr} (J/m²s), area of the shear plane, A , and the material removal rate, R (m³/s) [12],

$$U_s = \frac{U_{sr} A}{R}. \quad (10)$$

Specific shear energy rate is the shear energy rate generated per unit area and can be calculated once the area of the shear (deformation) zone is known, assuming that the shear is occurring on a plane [12]. Material removal rate (R) is defined as:

$$R = V t_0 t_w.$$

The area of the shear plane is:

$$A = \frac{t_0 t_w}{\sin \phi} \quad (11)$$

where ϕ is the shear plane angle (Fig. 2.6). In order to project the uncut chip area to the area of the shear plane, $\sin \phi$ is taken. Furthermore, the shear plane angle ϕ is given in terms of the tool rake angle and the chip thickness ratio :

$$\phi = \tan^{-1} \left(\frac{\frac{1}{\lambda} \cos \alpha}{1 - \frac{1}{\lambda} \sin \alpha} \right). \quad (12)$$

At a given instant, the shear flow stress on the shear plane (τ) is the shear force (F_s) (Fig. 2.6) divided by the area of the shear plane (A):

$$\tau = \frac{F_s \sin \phi}{t_0 t_w}. \quad (13)$$

Therefore the specific shear energy rate becomes:

$$U_{sr} = \frac{F_s V_s \sin \phi}{t_0 t_w} \quad (14)$$

where V_s is the shear velocity on the shear plane and can be given as $V_s = \gamma V \sin \phi$. Then, by combining Equations 9, 10 and 14, temperature rise in the chip can be rewritten as:

$$\Delta T = \frac{\beta_1}{\rho C_p} \frac{F_s V_s \sin \phi}{t_0 t_w} \frac{1}{\sin \phi V t_0 t_w} = \frac{\beta_1}{\rho C_p} \frac{F_s V_s}{V t_0 t_w} = \frac{\beta_1}{\rho C_p} \frac{\gamma F_s \sin \phi}{t_0 t_w}. \quad (15)$$

Since;

$$\frac{V_s}{V} = \frac{\cos \alpha}{\cos(\phi - \alpha)} \quad (16)$$

$$\Delta T = \frac{\beta_1}{\rho C_p} \frac{\gamma F_s \sin \phi}{t_0 t_w} \equiv \frac{\beta_1}{\rho C_p} \frac{F_s \cos \alpha}{t_0 t_w \cos(\phi - \alpha)} \quad (17)$$

then the temperature in the chip at the steady-state conditions will be:

$$T = \Delta T + T_0 \quad (18)$$

where T_0 is the initial temperature of the workpiece. This model ignores any temperature rise through the friction between the cutting tool and the workpiece. Although there can be an extra heating due to this friction, it is assumed that this heating will create only a local temperature rise at the interface. This temperature rise will not be experienced by the whole body of the chip; hence it will not result in a homogeneous temperature rise in the entire deformation zone that is enough to change the plasticity of the workpiece. Since this work focuses on the temperature effects on the deformation zone, the chip temperature right after the deformation zone is assumed to be equal to that of deformation zone.

The infrared (IR) thermography technique is employed in free machining in order to measure the temperature rise in the deformation zone [25]. IR thermographs show that the primary deformation zone is the main contributor to the temperature rise in the workpiece ahead of the deformation zone and in the whole body of the chip.

As seen from the Eqs. 17 and 18 above, the total temperature rise in machining depends on two unknown parameters β_1 and F_s . In the following sections these two parameters are explained in detail according to the related machining conditions.

2.3.1 The Heat Partition Coefficient, β_1

As there are many models previously describing this heat partition successfully, the model of Weiner [26] will be presented here. This model is shown to be relatively accurate in predicting the temperature rise in machining as compared to the experimental data of Boothroyd [27]. Like the other models, this model depends strongly on the initial chip thickness (or depth of cut or feed rate in the case of orthogonal machining), the cutting velocity and the thermal diffusivity. Thermal diffusivity is defined as: $\Omega = k/\rho C_p$ (m^2/s); where k is the thermal conductivity in W/mK , ρ is the density in kg/m^3 and C_p is the specific heat capacity in J/kgK .

The heat partition coefficient also depends weakly on deformation geometry, through the rake angle and chip thickness ratio, which are described through shear plane angle ϕ .

Then for this model, $\beta_1 = 1 - \Gamma$ where Γ is given as [26]:

$$\Gamma = \frac{1}{4Y} \text{erf}(\sqrt{Y}) + (1 + Y) \text{erfc}(\sqrt{Y}) - \frac{e^{-Y}}{\sqrt{\pi}} \left(\frac{1}{2\sqrt{Y}} + \sqrt{Y} \right) \quad (19)$$

with

$$Y = \frac{\rho C_p V t_0 \tan \phi}{4K} = \frac{V t_0 \tan \phi}{4\Omega}. \quad (20)$$

The change in the heat partition coefficient is depicted in Figure 2.5 for different machining velocities and thermal diffusivities. The coefficient is highly dependent on the thermal properties. As seen from the figure, it quickly saturates for a low thermal diffusivity and the conditions become more adiabatic as deformation rate approaches to 1 m/s.

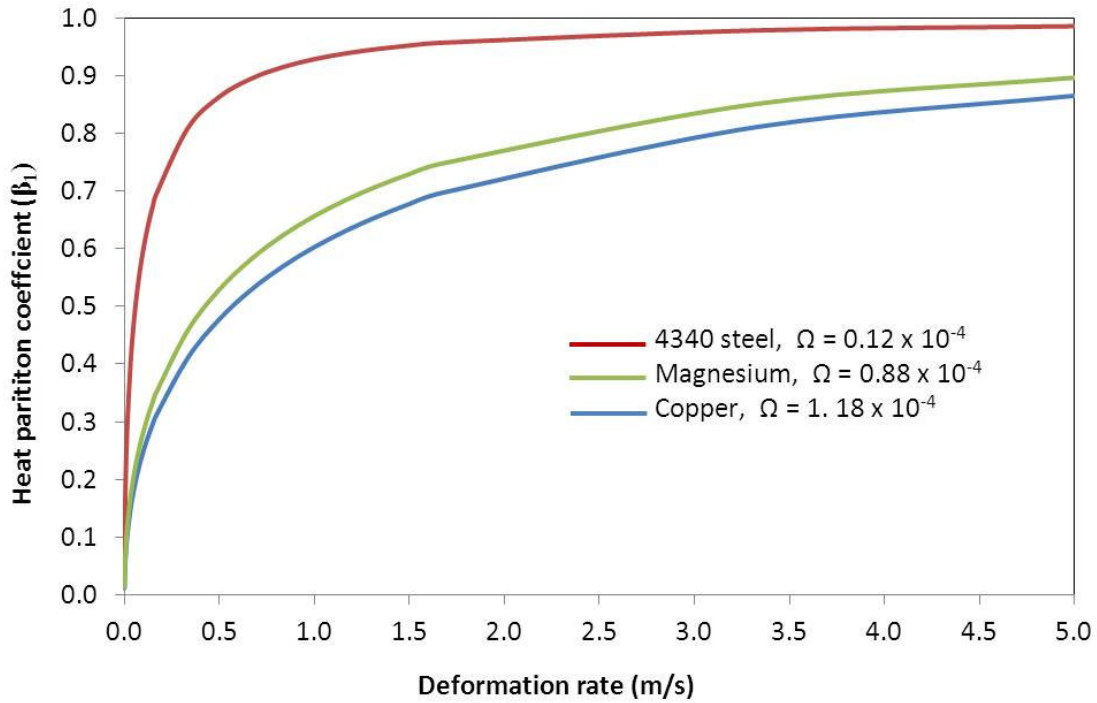


Figure 2-5 “Heat partition coefficient” (β_1) as a function of deformation rate (i.e. machining speed). Different thermal diffusivities (m^2/s) are compared.

2.3.2 Shear Energy and Cutting Forces

The forces acting on the chip in orthogonal free machining are given in the figure below (Fig. 2.6). As described above, the shear force is F_s and the normal force acting to it is F_n . The friction force between the cutting tool and the chip (workpiece) is F_f and the normal force acting to it is N .

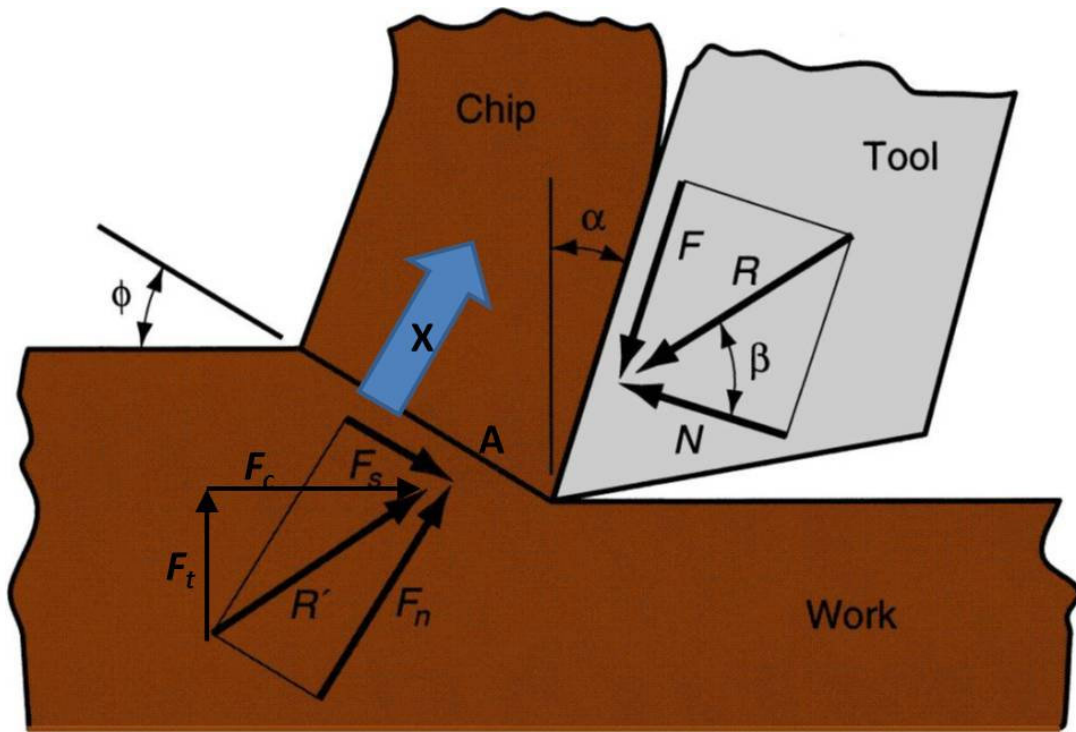


Figure 2-6 Forces acting on the chip in orthogonal free machining are given together with the measured forces (after [28], used with permission of Wiley).

The forces acting on the chip must be in balance; therefore resultant forces should be the same in magnitude, opposite in direction and collinear. However, none of these forces are measurable in machining. Instead measured cutting forces are often reported as the cutting force F_c and the thrust force F_t . Figure 2.6 gives the relationship of these measured forces to the actual machining forces.

Then following equations are derived from the force balance :

$$F = F_c \sin \alpha + F_t \cos \alpha \quad (21)$$

$$N = F_c \cos \alpha - F_t \sin \alpha \quad (22)$$

$$F_s = F_c \cos \phi - F_t \sin \phi \quad (23)$$

$$F_N = F_c \sin \phi + F_t \cos \phi. \quad (24)$$

From these equations, the shear force can be calculated from measured forces and used in the temperature rise equation (Eq. 17). In previous literature, shear force has been modeled through the shear flow stress. Constitutive models such as Johnson-Cook [29] and Zerilli-Armstrong [30] have been used to calculate the shear flow stress at a given temperature, strain and strain rate. However, these models require extensive testing to determine material properties and constants to model the flow stress. By direct measurement of cutting forces, extensive modeling of the flow stress can be avoided and the shear force can be directly calculated. Then a set of machining velocity, rake angle and feed will result in a shear force for a corresponding strain, strain rate and temperature in the deformation zone. This force divided by the shear plane area will be equivalent to the shear flow stress given by the constitutive models.

Consequently, an accurate description of the temperature rise in the deformation zone can be given by calculating the heat partition coefficient (from Eqs. 19 and 20) for a set of processing parameters and measuring cutting forces with a dynamometer attached to the tool (Eq. 17). Indeed, this temperature rise model is tested by experimentally measuring the temperatures in the deformation zone by employing the infrared (IR) thermography technique [25]. A camera that is capable of measuring the temperatures was focused on the middle of the deformation zone in free machining of pure copper in order to obtain IR thermographs. Then, the Johnson-Cook model was used to find the flow stress at the given set of machining conditions. Then, the temperature rise was calculated by using equations 17, 19 and 20. The calculated temperature was very close (± 10 °C) to the measured temperature obtained through IR thermographs.

Figure 2.7 gives the temperature versus deformation rate (i.e. machining velocity) graph for various pure metals assuming a deformation rate-independent shear energy. The shear energy values used in Figure 2.7 are typical to the conventional

free machining process. The increase in deformation rate causes an increase only in the heat partition coefficient through Eqs. 19 and 20.

The assumption of rate-independent shear energy is going to be tested in Chapters 5 and 6. Relatively high deformation zone temperatures and associated deformation mechanisms such as dynamic recrystallization are predicted to be the reasons of the constant shear force. Counter effects of strain rate hardening and thermal softening balance the shear stress. If this assumption does not hold for a given material, the shear force needs to be measured for each cutting velocity. Since measuring forces for each and every velocity would be impractical, a nominal shear force can be assumed for a range of velocities. Then Eq. 17 should be updated by the nominal forces for the given ranges. An alternative approach is to empirically relate the flow stress and velocity based on the experimental trends. This empirical relation can be inserted in Eq. 17 and the plot should be updated again.

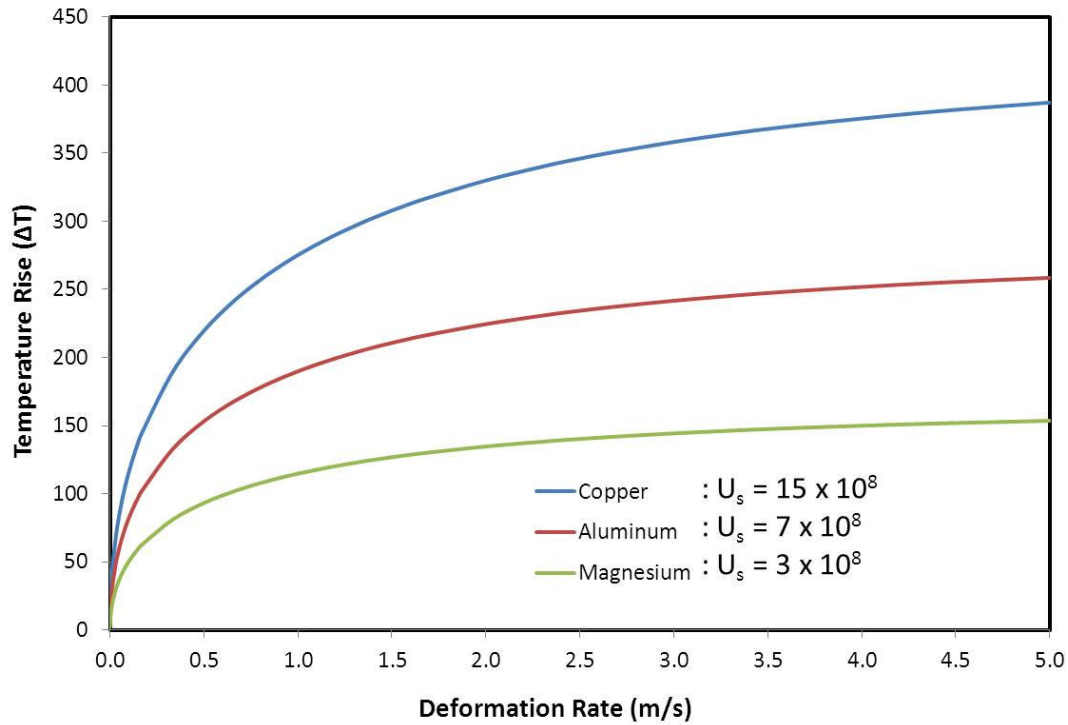


Figure 2-7 Temperature rise in the deformation zone for selected pure metals (Eq. 9). Note that these calculations are only representative and made by assuming deformation rate-independent specific shear energy (J/m³) based on force measurements. The higher specific energy of copper resulted in the highest temperature rise even though it has the highest thermal diffusivity among all these metals.

CHAPTER 3. SEGMENTATION IN MACHINING AND SHEAR LOCALIZATION MODELS

It is known that the main reason for segmentation and cracking in machining is the adiabatic shear banding (ASB) due to flow localization. The physics of shear banding is well understood and it is well known that susceptibility to shear banding strongly depends on the processing parameters. The effect of processing conditions on the resultant chip morphologies will be reviewed here in this section. Since all the literature in this area concentrates on free machining, the existing models need to be updated for the case of extrusion machining. The additional hydrostatic pressure term in EM needs to be taken into account.

Commonly observed chips in free machining are schematically depicted in Figure 3.1. The type (i.e. morphology) of the chip strongly depends on the type of the metal that is being cut and the given machining conditions.

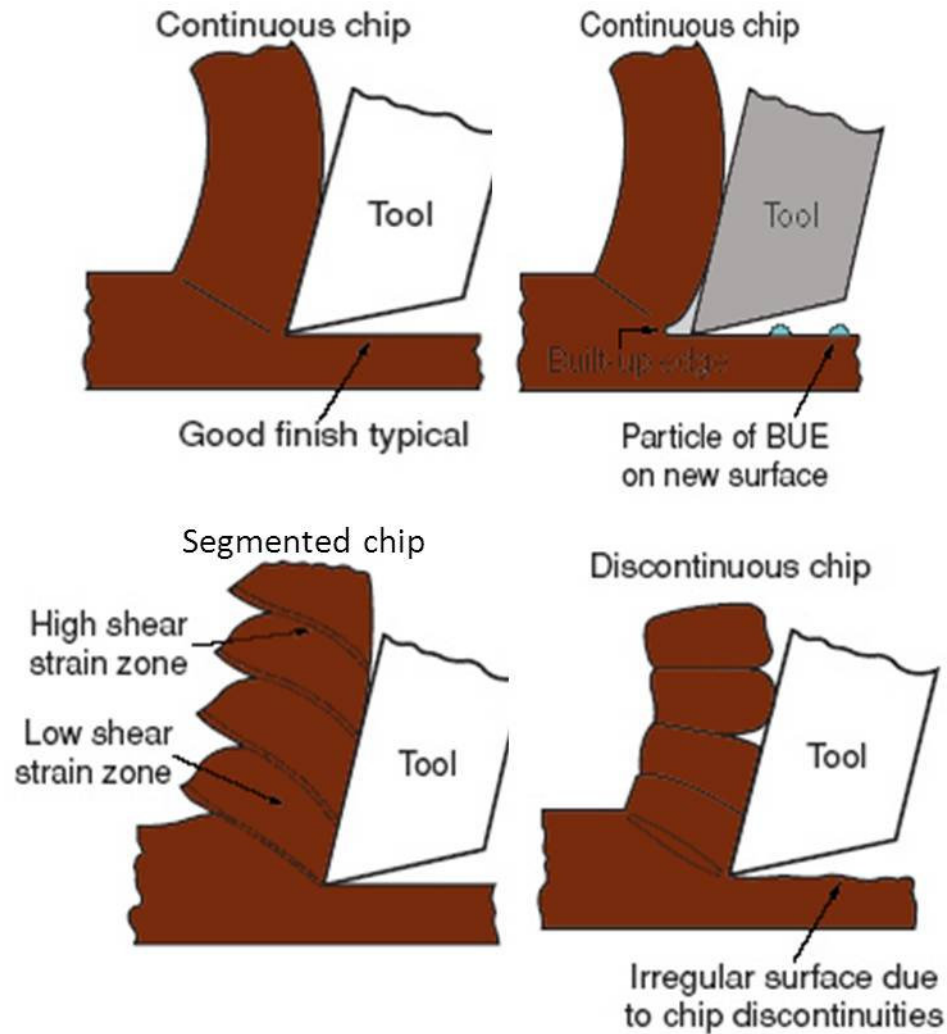


Figure 3-1 Common chip morphologies obtained during free orthogonal machining of metals and alloys (after [28], used with permission of Wiley). BUE = built up edge.

Continuous chips are usually a product of ductile metals at high velocities, small feeds, low friction between the tool and the chip and sharp cutting tools. Continuous chips with built-up edge occur also for ductile metals at low to medium velocities, low rake angles, high friction and blunt tools. These chips are usually detrimental for the desired surface finishes in both the bulk and the chip. Segmented chips occur in alloys with limited workability at high speeds and large feeds. They usually have periodic cracks located at the high strain regions. Materials that show minimal plasticity at room temperature result in discontinuous chips or powders at low

cutting speeds, large feeds and high friction. These chips are again unwanted for desired surface finish in the bulk, since they result in irregular surface due to discontinuities [28].

Both segmented chips and discontinuous powders can be classified in the same category. They both occur usually in alloys with limited workability and they have certain periodicity. Even discontinuous powders have very uniform size distribution because they tend to separate at the periodical cracks.

3.1 Load Instability Models

Earlier works on plastic instabilities started on tensile testing. Indeed, shear localization in machining is analogous to the necking in tensile testing. In both plastic instabilities, the stain and strain rate in the neck or the shear band differ significantly from the adjacent region where the deformation has stopped. Both the neck and the shear band are flow localized regions. The neck continues to deform once the localization starts and sustains further deformation where the shear band deforms to a limited amount and new periodical bands are generated to sustain further deformation. Thus, a simple load instability model which is first developed for tension is applied to the case of simple shear [31]. The load instability is initiated when load maximum is reached with the shear strain, $d\tau = 0$,

$$d\tau = \left(\frac{\partial\tau}{\partial\gamma}\right)_{\gamma,T} d\gamma + \left(\frac{\partial\tau}{\partial\dot{\gamma}}\right)_{\gamma,T} d\dot{\gamma} + \left(\frac{\partial\tau}{\partial T}\right)_{\gamma,\dot{\gamma}} dT = 0. \quad (25)$$

Assuming constant strain rate, $d\dot{\gamma}/d\gamma = 0$, heat generated equals to work done as in Eq. 8 for an adiabatic condition, i.e., $dT\rho C_p = \beta\tau d\gamma$, and a power-law constitutive equation for the stress where $\tau = B\dot{\gamma}^n$ (where B is a constant and n is the strain-hardening rate); the instability strain at the point of load instability $d\tau/d\gamma = 0$ can be derived as [31]:

$$\begin{aligned}
\left(\frac{\partial \tau}{\partial \gamma}\right)_{\gamma, T} &= -\left(\frac{\partial \tau}{\partial T}\right)_{\gamma, \dot{\gamma}} \frac{dT}{d\gamma} \\
Bn\gamma^{n-1} &= -\left(\frac{\partial \tau}{\partial T}\right)_{\gamma, \dot{\gamma}} \frac{\beta B \gamma^n}{\rho C_p} \\
\gamma_i &= -\frac{n\rho C_p}{\beta \left(\frac{\partial \tau}{\partial T}\right)_{\gamma, \dot{\gamma}}}.
\end{aligned} \tag{26}$$

From this equation, the susceptibility to strain localization or shear banding generally increases with low strain-hardening coefficient, n , large thermal softening coefficient $(\partial \tau / \partial T)$ and low thermal diffusivity through the increase in adiabatic conditions, β . Although the above equation can be used to predict susceptibility to shear banding in general, a more detailed analysis is needed for the present problem.

This theory does not address the details of nucleation and propagation of shear bands. It gives no information about their dimensions or shapes and ignores roles of imperfections and microstructure on the nucleation. It also neglects the stabilizing effects of strain-rate sensitivity. Indeed, experimental observations showed that instability strain is higher than the predicted value, especially for materials like nickel alloys, 4340 steel, titanium and tungsten [32].

A different approach was taken by Semiatin and Jonas [4, 33, 34] for the localization problem. This model maintains force equilibrium in the system. In order to maintain the force equilibrium, defects are introduced in the system. Therefore, differences in strain and strain rate that are required to maintain force equilibrium are shared between defective regions (shear bands in this case) and defect-free regions. Assuming plane strain deformation conditions, force transmitted to workpiece should remain constant along a specified direction at a given instant. By letting this direction (x) to be perpendicular to the cross section of the deformation plane (A) (Fig. 2.6), variation of the force along this direction can be written as:

$$\left(\frac{\partial \ln F}{\partial x}\right)_t = 0 = \left(\frac{\partial \ln A}{\partial x}\right)_t + \left(\frac{\partial \ln \sigma}{\partial x}\right)_t \quad (27)$$

where

$$d\sigma = \left(\frac{\partial \sigma}{\partial \varepsilon}\right)_{\varepsilon, T} d\varepsilon + \left(\frac{\partial \sigma}{\partial \dot{\varepsilon}}\right)_{\varepsilon, T} d\dot{\varepsilon} + \left(\frac{\partial \sigma}{\partial T}\right)_{\varepsilon, \dot{\varepsilon}} dT = 0. \quad (28)$$

Under plane strain conditions, the cross-section of the deformation plane does not vary with x (Fig. 2.6). Then in order to satisfy the above equation, flow stress should not vary with x as well:

$$\left(\frac{\partial \ln \sigma}{\partial x}\right)_t = 0 = n' \left(\frac{\partial \varepsilon}{\partial x}\right)_t + m \left(\frac{\partial \ln \dot{\varepsilon}}{\partial x}\right)_t. \quad (29)$$

Where n' is defined as a normalized, constant strain rate, strain-hardening rate and m is the usual strain-rate sensitivity. The familiar work hardening rate (n) which is measured under constant strain rate and temperature can be incorporated in the following equations as well:

$$n' = \frac{1}{\sigma} \left(\frac{\partial \sigma}{\partial \varepsilon}\right)_{\dot{\varepsilon}} = \frac{\left\{ \left(\frac{\partial \sigma}{\partial \varepsilon}\right)_{\varepsilon, T} d\varepsilon + \left(\frac{\partial \sigma}{\partial T}\right)_{\varepsilon, \dot{\varepsilon}} dT \right\}}{\sigma d\varepsilon} \quad (30)$$

$$m = \left(\frac{\dot{\varepsilon}}{\sigma}\right) \left(\frac{\partial \sigma}{\partial \dot{\varepsilon}}\right)_{\varepsilon, T} = \left(\frac{\partial \ln \sigma}{\partial \ln \dot{\varepsilon}}\right)_{\varepsilon, T} \quad (31)$$

$$n = \left(\frac{\partial \sigma}{\partial \varepsilon}\right)_{\varepsilon, T}. \quad (32)$$

Then Eq. 29 becomes the following and a flow localization parameter can be defined as;

$$0 = n' \delta \varepsilon + m \delta \ln \dot{\varepsilon} \quad (33)$$

$$\alpha_f \equiv \frac{\delta \ln \bar{\varepsilon}}{\delta \varepsilon} = -\frac{n'}{m}. \quad (34)$$

In theory, flow localization happens when $\alpha_f > 0$. However, experimentally it has been observed that significant localization starts only after $\alpha_f > 5$ [4]. Material properties have significant effects on the localization. From Eqs. 29 and 34, stabilizing effects of strain hardening and strain-rate sensitivity can be seen. This stabilization is again analogous to the neck formation in the tensile testing. The onset of neck formation is usually delayed with increasing strain-hardening rate. The onset of the neck is again delayed by increasing strain-rate sensitivity, where material in the neck flows at a higher stress due to the increased strain rate in the neck and the material outside the neck region can catch-up.

The main reason of flow localization under adiabatic conditions is then the flow softening of the material with increasing temperature. The more the flow softening, the more will be the tendency to form adiabatic shear bands. Hence, materials whose flow stress tends to decrease significantly with temperature under constant strain and strain rate are more prone to localization.

In the case of orthogonal machining, while assuming plane strain and simple shear conditions, a similar flow localization parameter can be defined using the same criteria (Eq. 34) [4]. Assuming that the material is isotropic and von Mises criterion is applied so that $\tau = \bar{\sigma}/\sqrt{3}$, $\gamma = \sqrt{3}\bar{\epsilon}$ and $\dot{\gamma} = \sqrt{3}\bar{\dot{\epsilon}}$, where $(\bar{\sigma}, \bar{\epsilon}, \bar{\dot{\epsilon}})$ are effective flow stress, strain and strain rate, respectively;

$$\alpha_{f,\tau} \equiv \frac{\delta \ln \dot{\gamma}}{\delta \gamma} = \frac{1}{\dot{\gamma}} \frac{d\dot{\gamma}}{d\gamma} = \frac{\alpha_f}{\sqrt{3}} = -\frac{n'_\tau}{m} \quad (35)$$

where

$$n'_\tau \equiv \frac{\partial \ln \tau}{\partial \gamma} \quad (36)$$

$$m = \left(\frac{\partial \ln \sigma}{\partial \ln \dot{\epsilon}} \right)_{\epsilon, T} \equiv \left(\frac{\partial \ln \tau}{\partial \ln \dot{\gamma}} \right)_{\gamma, T}. \quad (37)$$

Then the flow localization parameter for orthogonal machining becomes,

$$\alpha_f = -\frac{\sqrt{3}}{m} \left\{ \frac{1}{\tau} \left(\frac{\partial \tau}{\partial \gamma} \right)_{\gamma, T} + \left(\frac{\partial \tau}{\partial T} \right)_{\gamma, \dot{\gamma}} \left(\frac{1}{\tau} \frac{dT}{d\gamma} \right) \right\} \quad (38)$$

which is expressed in all measurable material properties from standard tests except the term $(1/\tau)(dT/d\gamma)$. The term $(1/\tau)(dT/d\gamma)$ can be derived from the Equations 13 and 17 that relate temperature rise in the shear plane to the process parameters [12]. Then,

$$\frac{dT}{d\gamma} = \frac{\beta_1}{\rho C_p} \frac{F_s \sin \phi}{t_0 t_w} = \frac{\beta_1 \tau}{\rho C_p} \quad (39)$$

$$\left(\frac{1}{\tau} \frac{dT}{d\gamma} \right) = \frac{\beta_1}{\rho C_p}. \quad (40)$$

According to the theory, the more heat that stays in the deformation zone (i.e. increasing adiabaticity) the more will be the tendency for localization or in this case formation of adiabatic shear bands. The flow softening term is multiplied by the heat partition coefficient, which increases with machining speed and the feed rate. Unless the flow softening is stabilized by high strain-rate sensitivity or work hardening rate, segmented chips will be expected for higher processing speeds or feed rates.

3.2 Perturbation Analyses

A different modeling approach has been pursued by various researchers. These efforts can be tracked back to earlier works of linear perturbation analysis of Clifton and Bai [35, 36]. By this approach it was possible to calculate shear band thickness and spacing as well, which were not calculable by previous models. The assumptions for the models were: negligible elastic strains, adiabatic conditions, a constitutive relationship for flow stress ($\tau \approx \gamma^n \dot{\gamma}^m T^{-\nu}$, where ν is the thermal softening rate) which is not affected by strain rate history and simple shear. Gradients in strain and strain rate are assumed to occur in previously defined x

direction (in Fig. 2.6) (the direction perpendicular to the shear plane cross-section) as well. Then defining the equations that govern the motion, a linear perturbation analysis can be performed [35]. Oscillatory perturbations (periodic fluctuations) in x direction are introduced after the stress maximum and a perturbation will grow with time when instability (or localization) happens. The instability should occur in order to conserve momentum and energy, which govern the motion. In other words, oscillatory perturbations in x direction becomes stable when strain localizes in narrow periodic regions. These regions can be called defects as well, where strain rate and strain differs from the matrix significantly. This localization is identical to the adiabatic shear bands defined before. The equations governing the motion can be defined as conservation of momentum:

$$\rho \frac{\partial^2 \gamma}{\partial t^2} = \frac{\partial^2 \tau}{\partial x^2} \quad (41)$$

And conservation of energy:

$$\tau \frac{\partial \gamma}{\partial t} = \rho C_p \frac{\partial T}{\partial t} - k \frac{\partial^2 T}{\partial x^2} \quad (42)$$

where k is the thermal conductivity. Therefore, by the application of perturbation analysis for adiabatic conditions (i.e., $k = 0$), the condition for instability can be found:

$$\frac{-\tau_0 \left(\frac{\partial \tau}{\partial T} \right)_0}{\rho C_p \left(\frac{\partial \tau}{\partial \gamma} \right)_0} > 1. \quad (43)$$

This condition of instability is very similar to the condition of instability derived from the load instability model (Eq. 26). Clifton also achieved a similar condition for instability in the absence of heat conduction (i.e. adiabatic conditions). Therefore, initiation of instability starts right after the stress maximum ($\partial \tau / \partial \gamma = 0$) and does not depend on the strain-rate sensitivity. Perturbations become stable when

thermal softening overcomes strain hardening. However, these models –both load instability and linear perturbation analysis– can predict only the onset of localization. Onset of the instability does not necessarily mean growth of perturbations will be catastrophic. Flow can localize after the instability point is reached, however this localization may not be catastrophic to form shear bands and segmentations where strain in the shear band is significantly more than in bulk [32]. Indeed, a critical time can be found in the work of Bai which is analogous to reciprocal of flow localization parameter [35]. When this critical time is reached, the perturbations reach a maximum extent; a point resembling the catastrophic shear banding. This critical time can also be treated as critical strain where flow localization reaches a maximum and shear banding occurs. Therefore, the reciprocal of this critical time or strain (α_m) is the flow localization (or susceptibility to shear banding) term of this perturbation analysis. Again assuming adiabatic conditions and a strain hardening metal, α_m becomes:

$$\alpha_m = -\frac{\sqrt{3}}{m} \left\{ \frac{1}{\tau} \left(\frac{\partial \tau}{\partial \gamma} \right)_{\gamma, T} + \frac{\beta_1}{\rho C_p} \left(\frac{\partial \tau}{\partial T} \right)_{\gamma, \dot{\gamma}} \right\} \quad (44)$$

which is similar to the one found by the load equilibrium criteria. This parameter should be greater than 0 as in the previous case to satisfy the conditions for shear banding. The stabilizing effect of strain-rate sensitivity is included in the flow localization parameter. This means, even after the point of instability the highly localized shear bands can be avoided by high strain-rate sensitivity.

Linear perturbation analysis is essentially a nonlinear problem. The coefficients in the linear differential equations were assumed to change slowly. Neglected effects of the time dependence of the coefficients and the nonlinearity of the equations are the problems of linear perturbation analysis [32]. Molinari and co-workers improved the theory by developing a relative linear perturbation analysis [37, 38]. Then, a condition for catastrophic localization can be derived directly:

$$m + \frac{1}{\tau} \left(\frac{\partial \tau}{\partial \gamma} \right)_{\gamma, T} + a < 0 \quad (45)$$

$$\text{with } a \equiv \frac{\beta_1}{\rho C_p} \left(\frac{\partial \tau}{\partial T} \right)_{\gamma, \dot{\gamma}}. \quad (46)$$

This conditions is again very similar to the ones derived before if a is the thermal softening parameter. For a perfectly plastic material or no strain hardening ($n = 0$), the term $(1/\tau)(\partial\tau/\partial\gamma)_{\gamma, T}$ will be 0 in all of the above equations. Then, shear banding is a competition between thermal softening and strain-rate sensitivity only; the former increasing susceptibility and the latter stabilizing the flow and decreasing susceptibility.

Wright and co-workers developed a nonlinear theory and came up with similar susceptibility coefficients ($\chi_{SB} \equiv \alpha_f$ or α_m):

$$\chi_{SB} = -\frac{a}{m} \quad (47)$$

for non-strain hardening materials and

$$\chi_{SB} = -\frac{a\sqrt{m}}{n\sqrt{m} + m\sqrt{n}} \quad (48)$$

for strain hardening materials [39].

In the case of free machining, all of these flow localization parameters can be used to predict segmentation, since they are all similar in nature. The heat partition coefficient (β_1) can be used instead of β for the case of machining. Then, for constant strain-hardening rate and strain-rate sensitivity, segmentation is expected to increase with more thermal softening. Since the thermal softening term will increase with an increase in the heat partition coefficient, β_1 , any increase in process speed and feed rate should theoretically increase susceptibility to segmentation. These process parameters will be experimentally varied for alloys

with limited workability (Mg AZ31) and these theoretical suggestions will be tested in Section 6.2.5. Since several researchers derived essentially similar coefficients for susceptibility to shear banding through different models, a universal susceptibility to shear banding factor will be used.

A common assumption both in load instability and perturbation analyses models is the constant material properties over a range of thermo-mechanical processing conditions. In Section 6.2.5 of the thesis, this assumption will also be tested for the magnesium. Susceptibility to shear banding will be plotted with a constant flow softening rate and strain-rate sensitivity over a range of thermo-mechanical conditions that are relevant to machining. Then, this plot will be modified with variable thermal softening and strain-rate sensitivity that are functions of strain, strain rate and temperature.

3.3 Flow Localization in Extrusion Machining

Extrusion machining differs from the orthogonal free machining (peeling) in three ways. One aspect is increased cutting forces with the application of the constraining tool. Simultaneous extrusion and cutting demands higher cutting forces. Some of these forces will be resolved on the shear plane, which will increase the temperature rise in the shear zone.

Increase in forces does not necessarily affect the heat partition coefficient, β_1 . Therefore, increased forces do not have a direct effect on the thermal softening term in the flow localization equations, since they do not necessarily increase the adiabaticity. However, the total temperature in the shear zone will be higher. The flow stress, τ , will be higher since it has a direct relationship with the forces.

An increase in the deformation zone temperature may change the deformation mechanisms of a given material. Plastic behavior of the material and the terms governing plastic behavior such as flow stress, strain-hardening rate, thermal softening rate and strain-rate sensitivity may change. Therefore, when free

machining and extrusion machining are compared the additional temperature rise in EM should be accounted.

A second aspect is the increased hydrostatic pressure in the shear zone. Recently, increased slip system activity with the application of hydrostatic pressure was shown in magnesium [40]. Increased slip activity may also affect the above listed material properties. The localization equations may need to be modified for the difference in material properties.

According to the theory of plasticity, hydrostatic pressure should have no effect on the flow stress. However, experimental evidence in the literature suggests there can be effects. Based on the experimental results, Spitzig gives an empirical relation between flow stress and hydrostatic pressure for a general case in the following [41]:

$$\sigma = \sigma_0(1 + 3\eta p) \quad (49)$$

where σ_0 is the initial flow stress, η is the pressure coefficient that is measured experimentally and p is the pressure. The experimental evidence provided in the literature has recently raised some attention in theoreticians, and progress has been made on the incorporation of applied hydrostatic pressure in the theory of plasticity [42].

In this thesis, hydrostatic pressure is assumed to have no effects on the flow stress and the plastic behavior. Dodd and Atkins also showed that the hydrostatic pressure does not have any effect on the adiabatic shear banding behavior of the solids that does not contain voids [43]. However, one can also modify the shear banding equations with an additional term for the pressure effects. The hydrostatic pressure can be calculated in extrusion machining as shown previously. Then the partial derivative of the flow stress with hydrostatic pressure can be added as an extra term into Eq. 28. Then, force equilibrium needs to be satisfied again with this new

addition. The stabilizing effects of hydrostatic pressure should be obvious through the pressure hardening of flow stress shown by Eq. 49. It will be a competing factor against the flow softening.

The final aspect is the shape of the deformation zone in extrusion machining. It was shown that the deformation zone in EM is narrower and more confined compared to the FM [44]. Thickness of the shear zone, Δ , is smaller in EM. This may affect the shear banding behavior. A high hydrostatic pressure in a confined deformation zone may decrease the cracking tendency along shear bands. Crack growth along shear bands, assuming shear banding behavior does not change, will be harder in EM.

Lapovok et al. incorporated the finite thickness of shear zone in their perturbation analyses model for predicting the adiabatic shear banding behavior during ECAP of Mg AZ31 [45]. The thickness of the shear zone imposes a boundary condition to the possible wavelengths of perturbations. There is a limited number of wavelengths allowed in the shear zone and some perturbations may not have time to grow into the shear bands. In other words, with decreasing shear zone thickness, the critical time when perturbations grow to the maximum extent may increase. Susceptibility to shear banding then decreases with increasing critical time, since critical time is the reciprocal of the flow localization parameter (Eq. 44).

CHAPTER 4. A REVIEW ON THE DEFORMATION BEHAVIOR OF ALLOYS WITH LIMITED WORKABILITY

As discussed in the introduction section, the materials studied in this thesis are divided into two major categories. The first category is the materials with intrinsic susceptibility to cracking due to limited plasticity. Deformation mechanisms of these materials show a significant change with the temperature, as in the example of magnesium AZ31. Usually in addition to their limited plasticity, these alloys are also characterized by their low thermal diffusivity and high thermal softening rate. Consequently, under high strain rate deformation and adiabatic conditions, adiabatic shear banding is the major reason of cracking (segmentation) for this category of materials.

Due to the commercial significance and the recent interest in magnesium alloys, the deformation behavior and plasticity are reviewed in detail in the following Sections 4.1 and 4.2. These sections will also establish a background for the results and discussions in Chapter 6.

The second category is the materials subject to cracking due to extrinsic defects like porosity, weak second phases or non-uniform microstructure. Cast alloys are good example of this category. In this case, adiabatic shear banding is not necessarily the reason for flow localization and segmentations. Isothermal geometric softening due to pores or weak second phases can be the reason of segmentation [43]. If a material contains these defects (i.e. nucleation sites), voids or cracks can nucleate at a critical strain. As straining continues, plastic flow can localize between voids or cracks under isothermal conditions. Further straining in these localized regions may

eventually lead to thermo-plastic shear instability (adiabatic shear bands) which may cause the ultimate fracture through cracking. Hydrostatic pressure is shown to significantly change the localization behavior in these materials.

Optimum thermo-mechanical conditions will be tested and implemented for both classes of materials in order to avoid cracking due to either reason.

4.1 Materials with Intrinsic Susceptibility – Magnesium AZ31

Although magnesium is abundant in mineral form and can be refined and cast relatively inexpensively [46, 47], final product forms such as foils, sheets and strips can be quite expensive due to the poor formability of this HCP-structured metal [46-50], which is due to the limited number of slip systems at room temperature. Activation of non-basal slip requires heating magnesium and its alloys $\geq 200\text{ }^{\circ}\text{C}$ [10]. Even at this temperature, multiple deformation steps are needed, not only to reduce the thickness of the sheet, but also to refine its microstructure [49-51]. Sheets are traditionally formed from cast Mg ingots by either rolling or extruding into final shapes [50]; or the material is cast directly into sheet forms using twin-roll casting (TRC) [52, 53]. In the rolling (extrusion) of cast Mg ingots, workpiece and rolls (dies) are preheated to the high temperatures ($\sim 200\text{ }^{\circ}\text{C}$) needed to activate non-basal slip. Typically, 15 to 30% thickness reduction per pass is imposed in 3 to 7 stages, with intermediate annealing. Traditionally, processing speeds are also kept small ($\sim 10\text{ m/min}$) to avoid cracking [50]. These constraints result in large fabrication costs relative to those associated with production of alternative alloy sheets such as aluminum and steel. In TRC, roll speeds are again kept small in order to achieve uniform solidification and crack free sheets [50, 53]. The cast sheet then needs to be further processed by conventional rolling or differential speed rolling (DSR) [54, 55] in order to refine the cast microstructure and reduce the thickness below 2 mm, which is the minimum thickness achieved so far by TRC.

Together with DSR, different severe plastic deformation (SPD) techniques such as equal channel angular pressing (ECAP) [56-58], unequal channel angular pressing (UCAP) and high-pressure torsion (HPT) [59] also have been proposed to break-up initial cast microstructure and texture by intense shear deformation. Compared to conventional rolling, these techniques produce finer microstructures (1-3 μm grain size) and less intensive (0002) textures; which are essential for further sheet forming processes and superplastic forming [60, 61]. Typically, however, these techniques by themselves cannot be used to make sheet material directly from the cast billets, for they do not provide large shape changes. The high temperature deformation ($\sim 200\text{-}250\text{ }^{\circ}\text{C}$) is again essential through workpiece and tool pre-heating.

In rolling, there have been attempts to exploit the heating arising from deformation in order to minimize the need for workpiece pre-heating, maximize the % reduction per pass and increase productivity. This is achieved by increasing roller speeds to as high as ~ 250 to 1000 m/min . While this concept has proven viable [62], the large deformation zones typical of rolling (or extrusion) and the heat lost into the correspondingly large tools make conversion of plastic energy into temperature rise less inefficient compared to EM [50, 51].

4.2 Deformation Mechanisms, Flow Softening and Shear Banding Behavior of Magnesium AZ31

Wrought forms of magnesium, especially the sheet, can find great variety of applications in discrete products if production cost is reduced and formability improved. Recently, great progress has been made in understating the deformation behavior of magnesium. Several thermo-mechanical routes are proposed for increased formability and decreased cost. Consensus of the recent progress is the need of non-basal textures (i.e. tilted basal textures) with fine grain microstructures for optimum formability and strength. Multi-step and shear-type deformation is a convenient and well-proven way to achieve non-basal textures in the wrought forms

either by using one of the following processes or combining them: equal channel angular pressing (ECAP), differential speed rolling (DSR), multi-axis rolling.

These processes are usually performed at high temperatures ($> 200\text{ }^{\circ}\text{C}$) and low strain rates ($< 10/\text{s}$) in order to get continuous and defect-free forms, since magnesium is susceptible to shear localization and cracking at low temperatures. Although these conditions are successful in getting continuous forms with uniform microstructures, they might be inefficient and expensive for industrial scale production. Ideally, commercial production demands minimum number of processing steps, high speeds (strain rates) and low temperatures for cost and energy efficiency.

Moreover, the shear localization behavior of magnesium with respect to deformation mechanisms is poorly understood [45, 63-65]. Adiabatic shear banding due to flow softening is usually the reason for shear localization. In magnesium, the flow behavior and deformation mechanisms are highly dependent on the thermo-mechanical conditions. There is a marked increase in plasticity at high temperatures and the deformation mechanism switches from twinning to basal-slip and eventually to non-basal slip with increasing temperature. Due to the HCP structure, active slip is limited to the basal planes (0002) at low temperatures and low strain rates. The critical resolved shear stress (CRSS) for basal slip is almost 100 times lower than the non-basal (i.e. prismatic or pyramidal) slip [66]. However, as temperature increases, the CRSS of basal and non-basal slip becomes almost equal and plasticity considerably increases [50]. Barnett, compares the CRSS for various deformation mechanisms with respect to temperature (Figure 4.1) [66]. There is a sharp decrease of CRSS of non-basal systems above $200\text{ }^{\circ}\text{C}$. Therefore, twinning and basal slip are expected to contribute to the plasticity at low temperatures and strain rates. Twinning is especially predominant in the cases where basal slip is completely restricted due to the geometrical orientations (i.e. c-axis is oriented parallel to the loading direction). Cross slip to non-basal planes are also observed at

low temperatures depending on the grain size. As the grain size gets smaller the stress near grain boundaries becomes high enough to exceed the CRSS required for non-basal slip. The CRSS for twinning increases as grain size decreases ($< 10 \mu\text{m}$), favors the slip dominated flow and diminishes the twinning activity. Increasing grain size has detrimental effects on the ductility of magnesium at low temperatures due to the twin dominated flow. Cracks can easily nucleate and propagate along twin boundaries and may cause embrittlement [50].

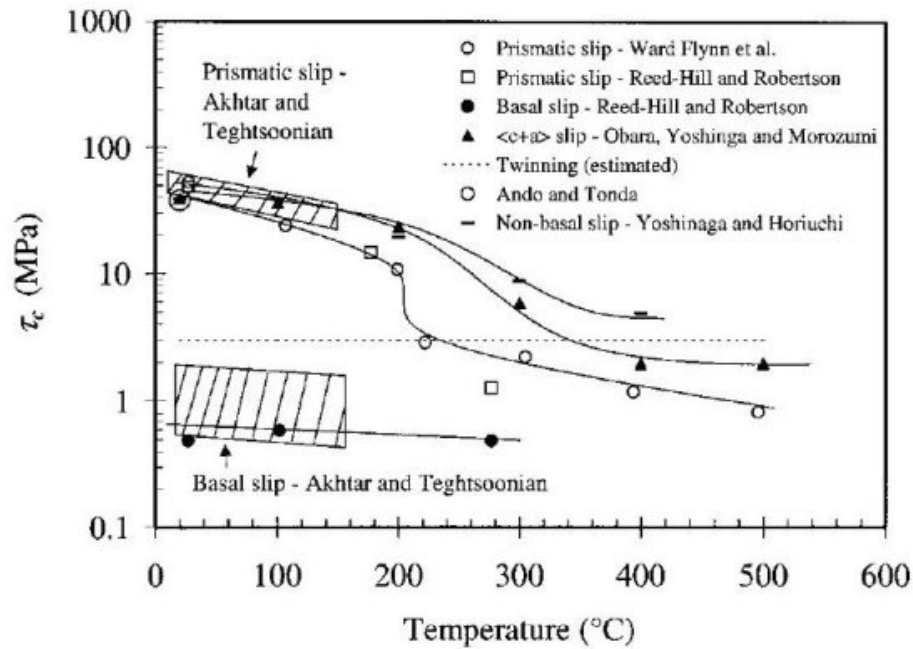


Figure 4-1 CRSS for various deformation mechanisms in pure magnesium with respect to the temperature at low strain rates ($< 10/s$) (after [66], used with permission of Springer).

Twinning almost completely diminishes at high temperatures and slip (both basal and non-basal) becomes dominant. Flow also becomes rate dependent and the rate has an opposite effect to that of temperature. Therefore, the flow behavior at high temperatures ($> 150 \text{ }^\circ\text{C}$) and low to moderate strain rates ($10^{-4} - 1/s$) can be modeled and explained by using an Arrhenius type constitutive equations [57, 67]:

$$\dot{\varepsilon} = A \left(\frac{\sigma}{E} \right)^n \left(\frac{b}{d} \right)^p D_0 \exp \left(\frac{-Q}{RT} \right) \quad (50)$$

where A is a constant, E is the elastic modulus, $n = 1/m$ is the stress exponent (it should not be mixed with strain-hardening rate), D_0 is the diffusion constant, b is the Burgers vector and Q is the activation energy for the diffusion (kJ/mol). The stress exponent, n, the grain size exponent, p and the activation energy, Q are the three most important variables that determine the hot working behavior.

Alternatively the Zener-Holoman parameter is used to understand the hot working behavior given that the activation energy is dependent and determined by the dominant deformation mechanism [68];

$$Z = \dot{\varepsilon} \times \exp \left(\frac{Q}{RT} \right). \quad (51)$$

At high Z values, (i.e. high strain rates – 0.1/s and high temperatures > 300°C) dislocation slip creep is the mechanism with dislocation climb controlling the rate and $n = 5-7$ and activation energy is equal to lattice diffusion. At low Z values, (i.e., low strain rates < 10⁻²/s and high temperatures) grain boundary sliding controls the deformation with $n = 2$ and activation energy is equal to lattice or grain boundary diffusion [50, 51, 57]. In Figures 4.2 and 4.3 the hot-deformation mechanisms is summarized in magnesium, together with an example deformation mechanism map drawn by using above variables and equation 50.

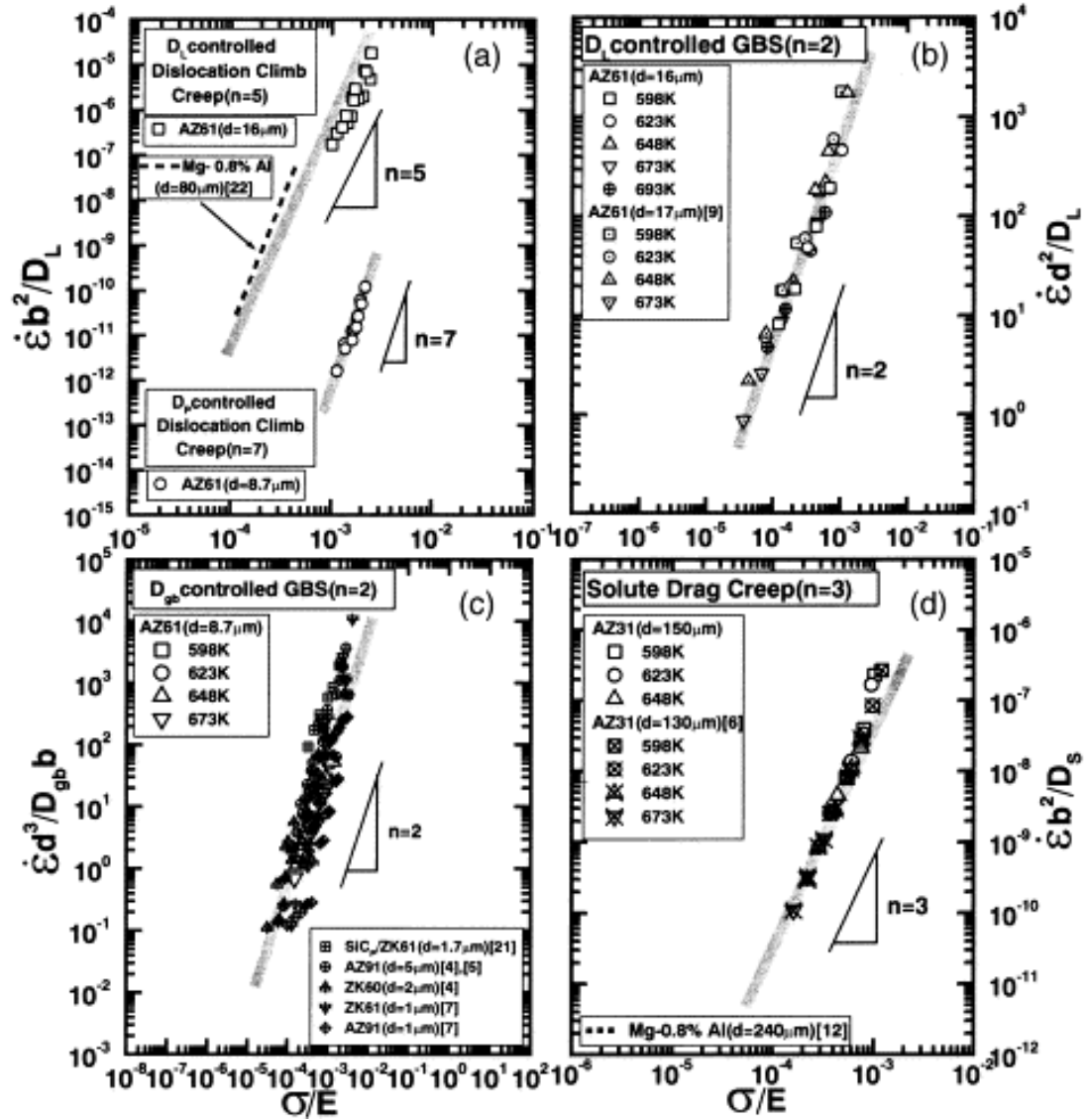


Figure 4-2 Diffusion-coefficient-compensated strain vs. elastic-modulus-compensated flow stress comparing the various deformation mechanisms and associated deformation variables at high temperatures (after [51], used with permission of Elsevier).

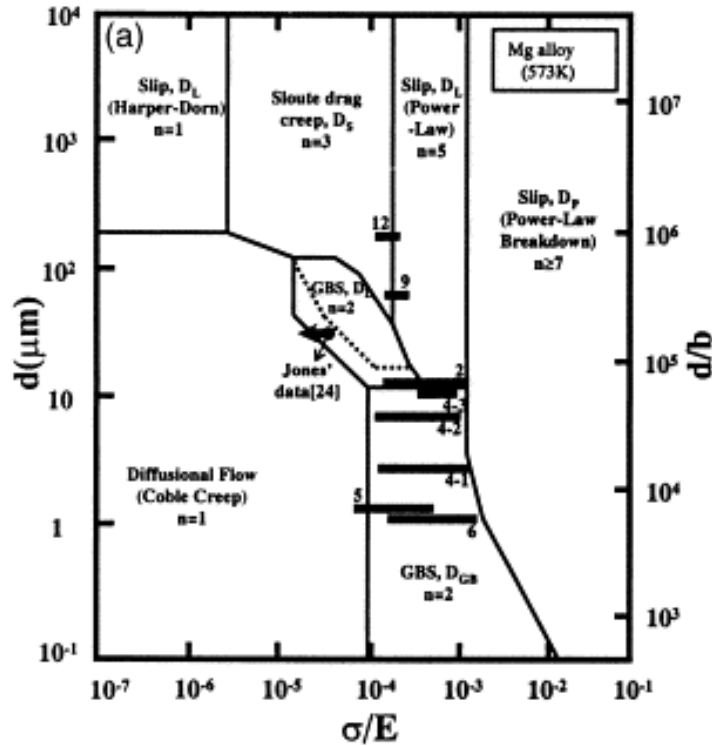


Figure 4-3 Deformation mechanism map of the magnesium alloy (after [51], used with permission of Elsevier).

It is important to note that the above given maps, equations and mechanisms are only valid over a limited thermo-mechanical range. The stress exponent and the activation energy will become temperature and strain rate dependent over a wider range and multiple mechanisms will be operative. A single rate controlling mechanism approach is not going to hold for high strain rate deformation. Moreover, Z cannot be correlated with deformation since information on Q is not available over extreme strain rates and low temperatures [69].

In order to model the adiabatic shear banding behavior, three material properties; flow softening rate, strain-rate sensitivity and strain-hardening rate need to be known or estimated. These properties cannot be assumed constant in magnesium, due to the marked difference between low temperature and high temperature plasticity and the number of various deformation mechanisms operating at these

domains. This remarkable switch in the plasticity is best depicted by the temperature dependency of strain-rate sensitivity (Figure 4.4). The strain-rate sensitivity increases 15-fold from $m \sim 0.01$ to 0.15 between room temperature and 200 °C, contributing extensively to the formability and ductility at high temperatures [70].

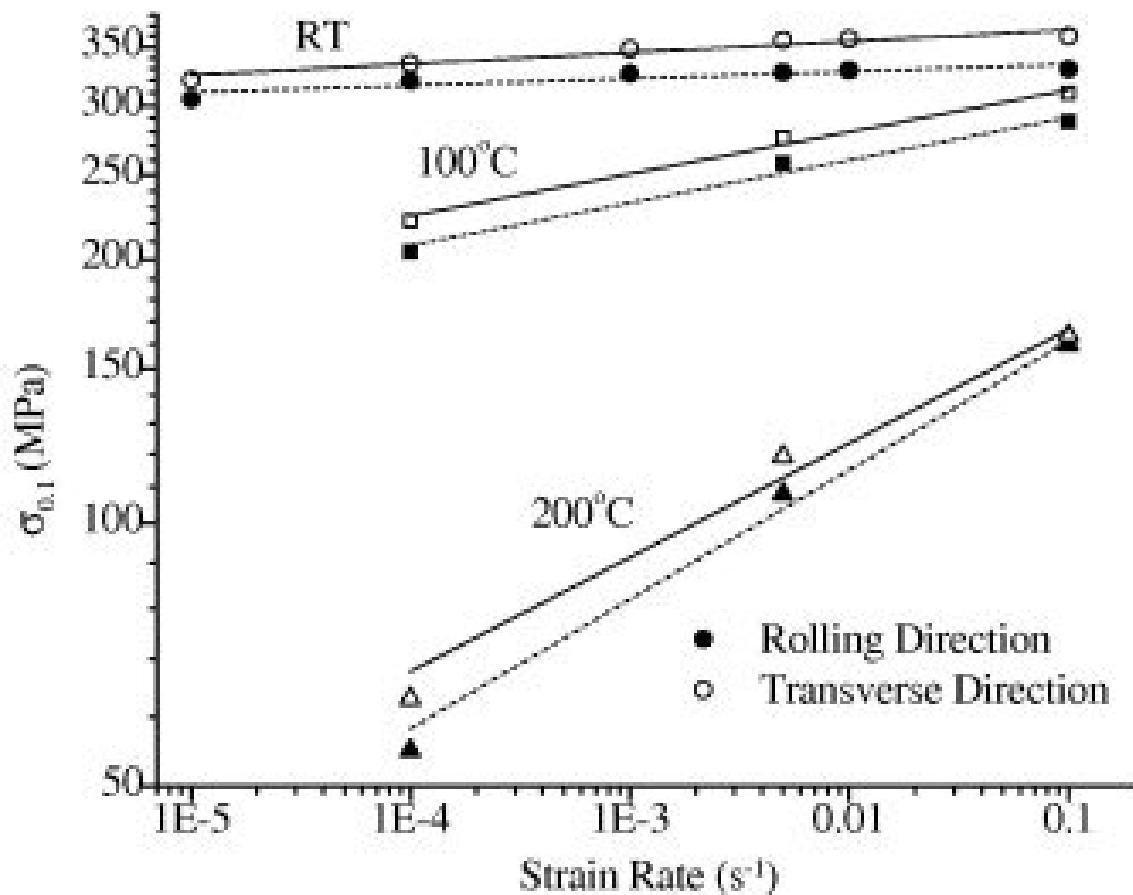


Figure 4-4 Strain-rate sensitivity vs. temperature in rolled Mg AZ31 sheet at $\varepsilon = 0.1$. TD and RD orientations are compared (after [70], used with permission of Elsevier). The typical texture in Mg AZ31 is the alignment of basal (0001) planes parallel to RD.

Temperature dependence of the strain-rate sensitivity is investigated by several researcher over a wide ranges of strains, strain rates and temperatures [67, 71-76]. A plot in Figure 4.5, summarizes the well-agreeing results of several researchers [67, 71-73, 75, 76]. A trendline is also fitted to the data and can be used as a guideline

when calculating m with dimensionless temperature which is defines as:
 $T = T(^{\circ}\text{C})/1(^{\circ}\text{C})$.

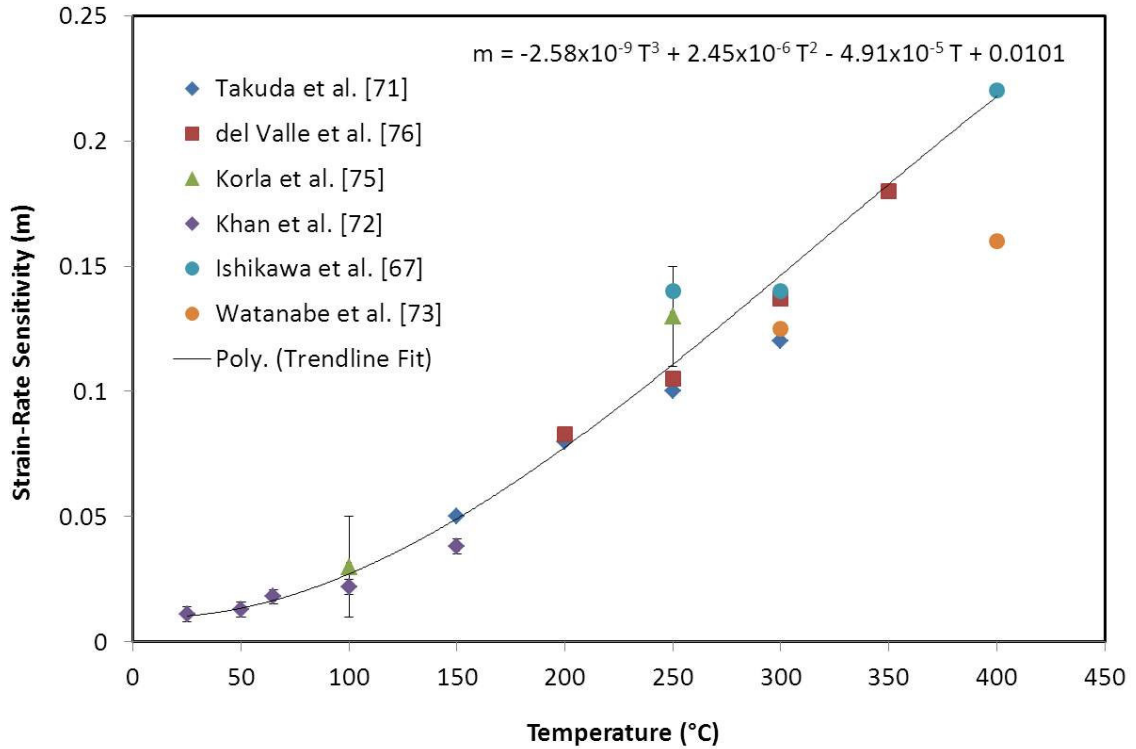


Figure 4-5 Strain-rate sensitivity (m) vs. temperature, valid for $10^{-3} \leq \dot{\epsilon} \leq 1/\text{s}$ and $0.05 \leq \epsilon \leq 0.7$.

While the sensitivity is little affected by the change in the strain rate and strain when $\epsilon > 0.1$ [71, 73, 75], the other properties such as temperature dependence of the flow stress (flow softening rate) and strain-hardening rate may vary. Thus, they need to be reported over a large range of strain rates and temperatures that are relevant to the shear banding conditions.

Strain-hardening rate (n) is almost temperature independent and is about 0.15 below 200 $^{\circ}\text{C}$ for strain rates bigger than $10^{-2}/\text{s}$. Above 200 $^{\circ}\text{C}$ it slightly decreases with temperature. At constant temperature, the trend is a 2 fold increase from ~ 0.1 to 0.2 with a strain rate increase from 10^{-3} to $10^{-1}/\text{s}$ [70, 71]. Then the hardening

rate saturates at the value of 0.2 with further increase in strain rate. It is fair to assume that, n changes in a narrow range of 0.1-0.2 over a wide range of strain rates and temperatures.

Temperature dependency of the flow stress over a wide range of strain-rates is given in Figure 4.6a-f [77]. Depending on the test type and orientation of the sample, usually a linear dependence is observed and a constant flow softening rate can be obtained for each strain rate. Flow softening rate decreases with increasing strain rate but still observed at $10^3/s$ [67, 77]. Elastic modulus is known to decrease with temperature as well. Modulus-compensated-flow-stress is plotted in Figure 4.7 with temperature [77]. This plot also is useful for finding the dominant deformation mechanism at a given temperature by using deformation mechanism maps (Figure 4.3).

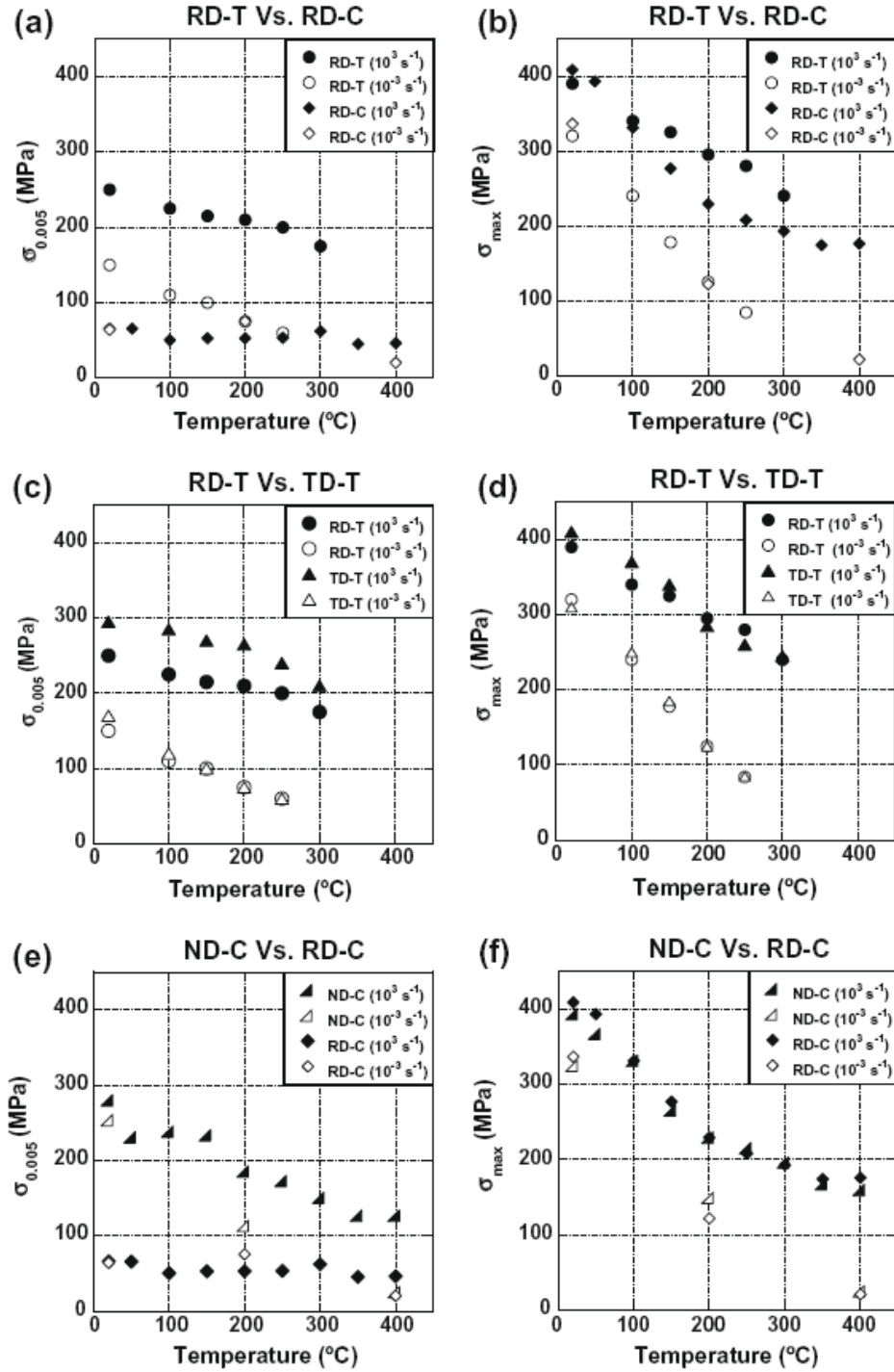


Figure 4-6 Tension (T) and compression (C) tests of magnesium AZ31 sheet (1 mm thick) at different strain rates, temperatures and orientations. True stress at $\varepsilon = 0.005$ ($\sigma_{0.005}$) and flow stress (σ_{max}) vs. temperature (a-f). (after [77], used with permission of Elsevier)

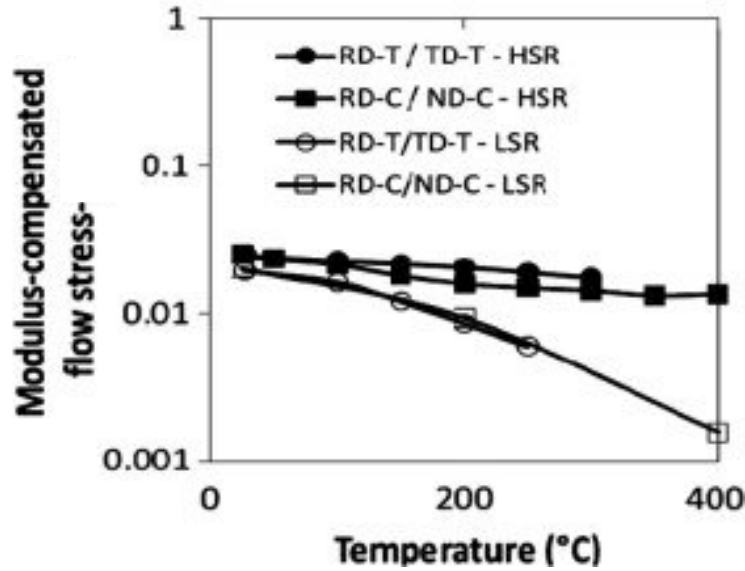


Figure 4-7 Modulus-compensated-flow-stress of the same magnesium sheet in Figure 4-6 vs. temperature at high strain rate (HSR) and low strain rate (LSR) (after [77], used with permission of Elsevier).

Modulus-compensated-flow stress becomes independent of temperature at high strain rates. Indeed, a constant value of $\sim 10^{-2}$ is obtained for σ/E , suggesting that a dislocation glide type deformation mechanism is dominant (Figure 4.3). This also indicates that a diffusion type process is not rate-controlling (Eq. 50) at high strain rates.

Dynamic recrystallization (DRX) takes place in magnesium simultaneously with deformation with the help of slip activity when temperature is above than 200 °C [10, 76, 78, 79]. The activation of DRX enhances ductility and the initial microstructures can be refined into smaller scale with possible textures that are different that the original [50]. The dynamic recrystallization is well understood as a function of temperature and strain [78], yet high strain rate studies are limited. There are mainly two types of dynamic recrystallization: continuous and discontinuous. Discontinuous dynamic recrystallization (DDRX) is the nucleation and growth of new grains in order to relieve the stored deformation energy during hot working and it is the preferred mechanism in materials with low to medium

stacking fault energy. Although magnesium is considered to have low stacking fault energy, DDRX is rarely observed in magnesium at warm deformation conditions (250 °C – 350 °C). As temperature increases and diffusion processes such as dislocation climb controls the deformation, DDRX becomes easier [50, 80]. The original grain boundaries serrate and new grains are formed by bulging of these serration. A necklace structure of recrystallized grains is first observed around original boundaries before all the original grains are consumed by small grains with increasing strain [10]. DDRX grains usually possess basal textures as in the case of hot rolling and they keep the texture of the original grains that they initiated from. However, a texture randomization can be possible with DDRX if the original grains have random texture distribution as well [10].

Continuous dynamic recrystallization (CDRX) or dynamic recovery happens when the stacking fault energy is high and cross-slip is possible. Cross-slip can be activated in magnesium at warm temperatures (250 °C – 350 °C) given that the grain size is small enough to allow accumulation of basal dislocations at grain boundaries where cross-slip to non-basal planes is easier. The stacking fault energy is considerably higher on non-basal planes, which allows further cross-slip and dislocation rearrangement. Dislocation cell structures and low-angle subgrain boundaries form by the accumulation and rearrangement of dislocations. Then, these substructures rotate to accommodate the deformation and the dislocation density increases in the substructures, resulting in the gradual formation of high-angle grain boundaries [50, 76, 79-81]. Rotational dynamic recrystallization (RDRX) and twinning-induced dynamic recrystallization (TDRX) can be given examples to CDRX, although the former is the usual mechanism and the latter is only observed around twin boundaries given that the temperature and strain is low enough for the twinning activity. CDRX grains have a texture in which c-axes are oriented parallel to the deformation axis (i.e. direction of maximum elongation) [50, 82].

In order to achieve uniform microstructures and defect-free wrought forms with ultrafine to fine grained microstructures, the dynamic recrystallization and shear banding behavior need to be understood over a range of thermo-mechanical parameters such as strain rate and temperature. Then, the microstructure and texture can be engineered without any shear banding by the choice of optimum processing conditions.

Adiabatic shear banding (ASB) occurs once the flow softening is possible and a critical strain level needs to be passed. As a consequence, non-uniform microstructures develop where strain and dynamic recrystallization is localized to the shear bands, together with relatively strain-free and coarse-grained matrix. In the meantime, cracks usually initiate and grow through the bands.

It is well known that, DRX occurs inside the adiabatic shear bands as a result of the high strain and temperature localized to the band. This causes the inhomogeneity in the microstructure and the shear band itself becomes brittle and prone to crack initiation and growth [83]. Recently some researchers claim that in magnesium and Ti-6Al-4V alloys, DRX occurs before the adiabatic shear band formation and causes perturbations in the microstructure and flow softening. DRX was suggested to be the main reason for shear band formation due to the perturbations in the microstructure. On the other hand, the classical theory of the shear banding indicates the changes in flow properties as the main reason [45, 84, 85]. While DRX is known to result in flow softening [76, 78], in most cases it increases ductility and workability of magnesium alloys by increasing the strain-rate sensitivity. Uniform microstructures are obtained without any shear band formation [50, 63, 86, 87]. Therefore, in magnesium alloys the shear banding behavior may be related more to the type of deformation mechanism and the type of DRX (i.e., TDRX, RDRX or CDRX) that is operating under a given range of thermo-mechanical conditions. In order to understand the shear banding behavior in magnesium, it is important to correlate

the instantaneous flow properties, the deformation mechanisms and the DRX at a given thermo-mechanical state.

Therefore, processing maps are needed to be established for the shear-type deformation processes at high strain rate and low temperature domains that are relevant to cost and energy efficient production and ASB formation. The extrusion machining process, being a single step and “simple shear” deformation process, becomes an ideal fit for modeling the previously proposed processes at large strains and strain rates and various temperatures.

CHAPTER 5. EXPERIMENTAL PROCEDURES

The methods for free and extrusion machining are described in the first section (5.1). These methods will be similar for all the materials tested. The materials section (5.2) mainly features the magnesium study. Techniques and results for all other materials are presented in a short section (5.3) after the magnesium study.

5.1 Experimental Parameters for Processing & Modeling

The processing parameters that govern the thermo-mechanical conditions of machining were introduced before. Rake angle and chip thickness ratio change the shear strain. Chip thickness ratio is not controllable in free machining, in contrast to the extrusion machining; hence it is a natural outcome of the process (Fig. 1.1). Deformation zone temperature is controlled by the initial temperature, heat partition coefficient and specific shear energy. Heat partition coefficient is heavily affected by velocity and feed rate; all other parameters are assumed to have minor effects. Specific shear energy depends on the initial dimensions of the chip, rake angle and cutting forces. Since the cutting force itself can be changed by initial temperature, velocity, feed rate, rake angle and chip thickness ratio, it will be measured for each of these parameters. Hydrostatic pressure depends only on the chip thickness ratio and rake angle. Although rake angle can be changed over a largerange it will be fixed for simplicity. The ranges of the variables investigated are given in Table 5.1.

Table 5-1 EM parameters controlling the thermo-mechanical conditions.

Thermo-mechanical condition	Process Parameter	Parameter Range
Effective Strain, ϵ	λ	0.5 to 3
Effective Strain Rate, $\dot{\epsilon}$	V, γ, Δ	0.02 to 4 m/s, 2 – 4, 50 to 100 μm
Deformation Zone Temperature, T	T_0, U_s, β_1	0 to 400 °C, given below, given below
Heat partition coefficient, β_1	V, f	0.02 to 4 m/s, 0.02 to 2 mm
Specific Shear energy, U_s	V, f, λ	0.02 to 4 m/s, 0.02 to 2 mm, 0.5 to 3
Hydrostatic Pressure, $p/2k$	λ	0.5 to 3

Force measurements for various conditions were performed with a Kistler dynamometer (Model No. 9257B) with the purpose of obtaining estimates of the specific shear energy. The tool was mounted on the dynamometer so that it directly measured F_c and F_t , as shown in Fig. 1.1. Forces were measured for a 20 second interval, which corresponds to at least 5 rotations of the workpiece. The measured cutting force was used to calculate the specific shear energy using the following equation and the corresponding temperature in the deformation zone;

$$U_s = \frac{2}{3} \frac{F_c}{t_0 t_w} \quad (52)$$

assuming 2/3 of total cutting force will resolve as F_s on the shear plane and the rest will overcome the friction. Eq. 52 is based on the established fact that, to first order, approximately 2/3 of the total energy of chip formation is dissipated at the shear plane [88].

In some of the experiments, the workpiece was preheated to a desired temperature. Workpiece temperatures were measured just before and immediately after the machining with a thermocouple contacted directly on the surface of the workpiece. The workpiece was mounted on a 7.5 cm diameter arbor during the pre-heating and

the machining. Typically, the assembly was heated in the Linderberg Blue M furnace for at least one hour to 50 – 100°C above the target temperature. All heating was done in open atmosphere. Workpiece and arbor was then chucked to the spindle and held until the assembly cools down to the desired pre-heating temperature. Heat losses to external environments are believed to kept minimum as the massive arbor helped to keep the heat inside the workpiece. The workpiece temperatures reported are accurate to within ~25 °C, which was the typical temperature difference between the measurement just before and right after machining.

The deformation zone temperature (T) was estimated using the shear plane model (section 2.3). The estimated T has been found to agree well with temperatures measured using infrared (IR) thermographic and thermocouple methods [22]. Hence, in the present study, T is taken as:

$$T = \frac{\beta_1 U_s}{\rho c} + T_0 \quad (53)$$

Equations 19 and 20 define the heat partition coefficient, β_1 . The time for which the material is exposed to T is quite small, at ~ 0.1 ms; this corresponds to the time taken for a material element to traverse the deformation zone (~ 100 μ m). The analysis enables control of T through appropriate selection of EM parameters.

5.2 Materials

5.2.1 Magnesium AZ31

MgAZ31B alloy tooling plate (hot-rolled condition) in disk form (ThyssenKrupp Materials NA) was subjected to plane strain machining as in Fig.1.1 (6.5 kW lathe). The disks were 6 to 25 mm in thickness and 120 mm in diameter. The material initial hardness was 58 ± 3 kg/mm² and grain size was 16 ± 2 μ m as shown in Figure 5.1. This picture was taken from the side surface of the disk, same orientation as the side view of the chip cross-sections. Texture data corresponding to the same

orientation is also given in Figure 5.1. Basal type texture that is common to wrought rolled forms was also observed for our as-received material.

Sheet created ranged from ~ 0.15 to 2 mm in thickness, with width equal to the thickness of the disks machined.

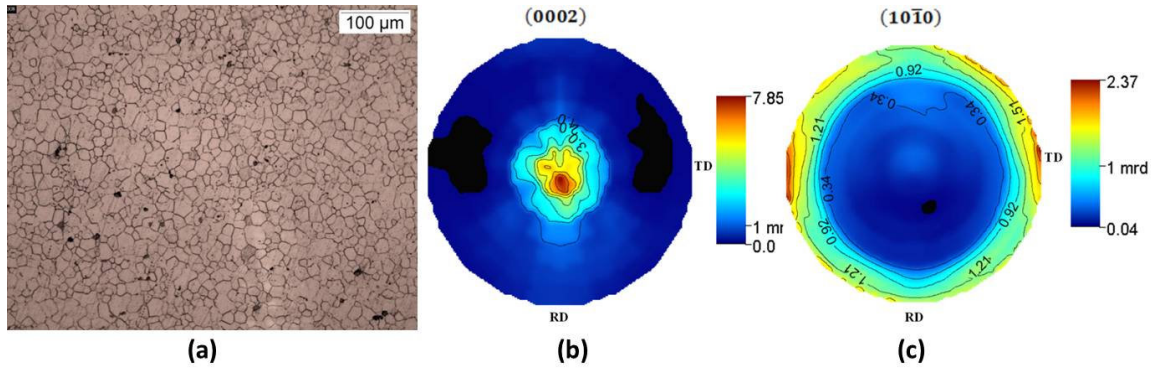


Figure 5-1 A typical optical micrograph of the wrought Mg AZ31. Black dots are precipitates of an intermetallic compound of magnesium and aluminum. Note the average grain size is in the order of 15 μm (a). (0002) and (10 $\bar{1}$ 0) pole figures are plotted for rolling and transverse directions of the plate (b & c). Basal type textures are common in wrought forms.

5.2.2 Experimental Conditions and Force Measurements for Magnesium AZ31

In all of the experiments, the rake angle (α) was kept constant at 5° . The shear strain was also kept at ~ 2 ($\bar{\epsilon} = 1.15$) by setting λ equal to 0.7 (high pressure, HP EM) or 1.4 (low pressure, LP EM). While the strain is the same at these two settings of λ , the deformation path and the hydrostatic pressure for the two cases are quite different (Fig. 2.3). Also, notably, for $\lambda = 0.7$, the thickness of the sheet exiting the deformation zone is smaller than to while for $\lambda = 1.4$, it is larger than to as in conventional machining.

The forces, F_c and F_t , were measured as described in section 5.1. The shear energy (U_s) was estimated from the forces measured for each process condition (Eq. 50).

Unless it is noted, the average value of U_s was $\sim 4.3 \times 10^8 \text{ J/m}^3$ and $2.3 \times 10^8 \text{ J/m}^3$ over the range of HP and LP EM conditions, respectively. Deformation temperature, T , was obtained using Eq (53). In some of the experiments, the workpiece was preheated, enabling T_0 to be set at up to $250 \text{ }^\circ\text{C}$ prior to the EM. This enabled additional control of T where needed, in particular to study (or eliminate) its influence on the deformation. T_0 was measured with a thermocouple just before the machining. The hydrostatic pressure was estimated using the SLF analysis (Fig. 2.3). Two levels of p were used, $p/2k = 1.7$ ($\lambda = 0.7$) referred to as high pressure EM (HP EM), and $p/2k = 1.1$ ($\lambda = 1.4$), low pressure EM (LP EM).

The variation of T with V_0 is shown in Fig. 5.2 for LP and HP EM, and for two values of T_0 , $25 \text{ }^\circ\text{C}$ and $200 \text{ }^\circ\text{C}$. T increases initially quite steeply with increasing V_0 before reaching a relatively constant value. This saturation value is greater in HP EM than in LP EM, and when T_0 is higher. The dependence of T on λ is analyzed likewise. The estimations enable T to be controlled by appropriate setting of λ , T_0 and V_0 (Table 5.2 and Fig. 5.2). Table 5.2 summarizes the deformation variables and the EM (input) parameters used to set the variables in the experiments.

Table 5-2 Measured cutting forces and specific shear energies for various processing parameters

Controlled variable	Range	EM parameter	Range
ϵ	2	λ	0.7 to 1.4
T	70 to $400 \text{ }^\circ\text{C}$	V_0, T_0, λ^*	0.02 to 2 m/s, 25 to $250 \text{ }^\circ\text{C}$, 0.7 to 1.4
$p/2k$	1.1 to 1.7	λ	0.7 to 1.4

* F_c , U_s and, hence, T are influenced by λ .

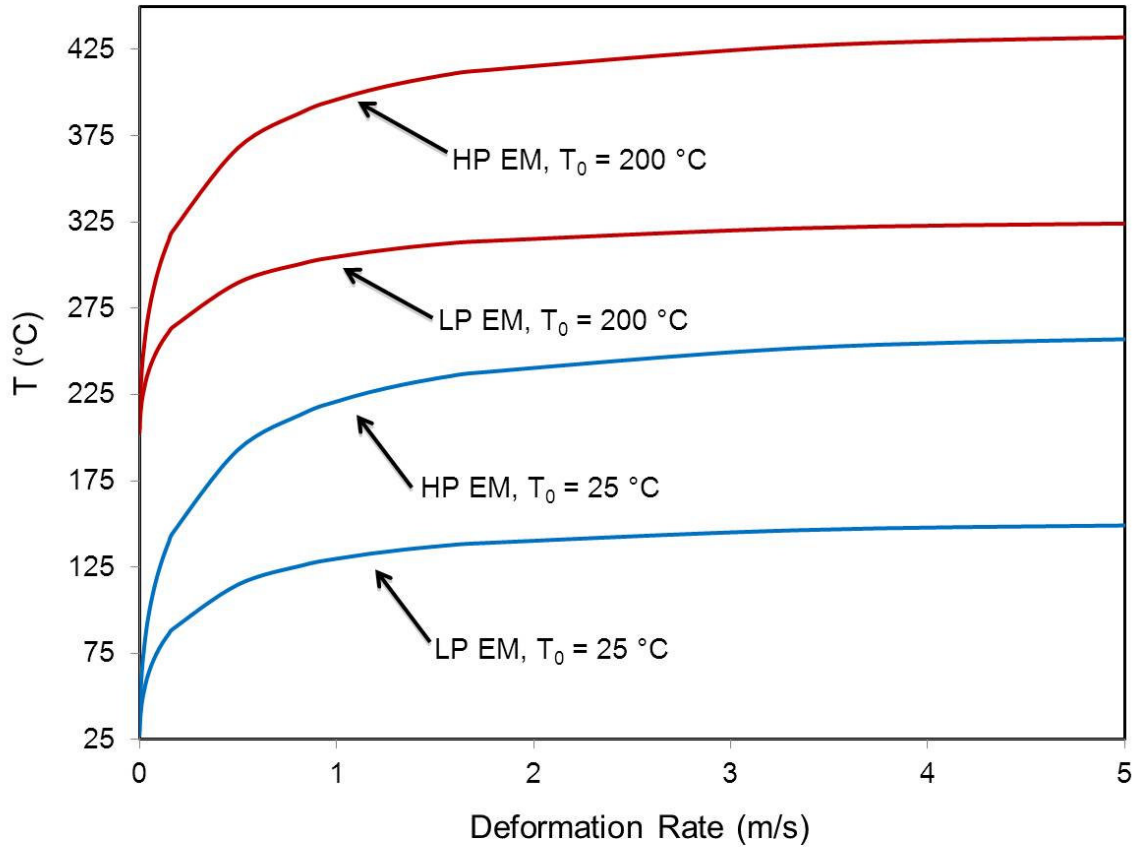


Figure 5-2 Variation of deformation temperature (T) with deformation rate (velocity) (V_0). For HP EM, $p/2k = 1.7$ and $U_s \sim 4.3 \times 10^8 \text{ J/m}^3$; and for LP EM, $p/2k = 1.1$ and $U_s \sim 2.3 \times 10^8 \text{ J/m}^3$ (after [11], used with permission of Elsevier).

5.2.3 Characterization of Magnesium AZ31

Grain size and strength properties of the sheet were characterized by optical microcopy, Vickers indentation, and tensile testing. Metallographic preparation was done by polishing with SiC abrasive papers of 300, 400, 600 and 800-grit, 3 and 1 μm diamond paste, and 0.05 μm colloidal silica, in that order. Ethanol was used as the polishing fluid with the SiC to minimize reactions with the Mg. Chemical etching for grain size analysis used a mixture of 5 ml acetic acid, 6 g picric acid, 10 ml water and 100 ml ethanol (200-proof). Grain size was measured with the Heyn linear intercept method, tracking at least 10 lines and 100+ grains in each sample. The

strip cross-section that is perpendicular to the chip flow direction was studied in optical microscopy and hardness testing.

Transmission electron microscopy (TEM) was used to characterize the UFG microstructures. TEM specimens were prepared by twin-jet electropolishing. Disk samples were punched from the sheet and electropolished directly without mechanical thinning. Electropolishing was performed between 30-50 V and 20-30 mA, using a solution of 1% perchloric acid – ethanol at -20 °C. Bright and dark field images were obtained using an FEI Tecnai microscope to estimate parameters such as grain size.

An area detector diffraction system (GADDS, from Bruker AXS, Madison, WI) with Cu K α radiation source operating at 40 kV and 40 mA was used to determine the quantitative crystallographic texture of all the samples. Unlike conventional scanning point detectors or position sensitive detectors, area (two-dimensional) detectors intersect the diffraction cones to form Debye arcs. This reduces the time needed to gather the diffraction data in two-dimensional space that is required for the quantitative texture analysis. The X-ray diffraction data was processed using a program, 'Material Analysis Using Diffraction' (MAUD) that employs Rietveld refinement technique for fitting the diffraction pattern. A more detailed explanation of the 2D x-ray method can be found in the previous work [89]. The strips were ground and polished from the surfaces until the mid-thickness of the strips is exposed and diffraction data was collected from a 0.5 mm diameter circle by focusing the x-ray beam with a collimator. This procedure allowed to sample a significant number of grains (crystallographic orientations) and to obtain diffraction data from the middle cross-section plane parallel to strip flow direction where interactions from original strip surfaces were avoided.

Electron back scattered diffraction (EBSD) technique was also employed to obtain micro texture and microstructure information. EBSD was also performed on the

same location as the x-ray diffraction. The strips were ground and polished again to the mid-thickness. The last polishing step was colloidal silica and this step was followed by electropolishing using a solution 37.5 vol% phosphoric acid and 62.5% ethanol (200-proof) at 4V and 5 °C for 45 minutes, since EBSD techniques is very sensitive to the final surface finish. A scanning electron microscope (SEM, FEI XL-40) equipped with an EBSD diffraction detector (EDAX) was employed with 30 kV acceleration voltage and 70° sample tilt. A specimen area 20 X 20 μm was scanned with a step size of 0.1 μm , which is typically 5 data point per grain. The manufacturer software (EDAX, OIM) was used for data analysis. The microstructures is presented by EBSD orientation maps and the texture pole figures were calculated form the orientation data. Grain boundaries having a misorientation angle greater than 15° are considered to be high-angle type and separate grains are defined by at least 5° misorientation.

Vickers indentation at 100 g indentation load (Leco Hardness Tester, model: LM247AT) was used to measure hardness (strength). At least 25 indents were made in a specimen to obtain mean values and standard deviations. Tensile test specimens of gauge length 10.4 mm were cut from the plane of the sheet with the length oriented parallel to the chip velocity (V_c , Fig. 1). The gauge section was 0.25 x 2 mm for the control specimens (as received tooling plate) and 0.18 x 2 mm for the EM sheet specimens. The tensile tests were done at 25 °C and a strain rate of $1.6 \times 10^{-2}/\text{s}$ on an MTS compression/tension test machine (MTS Insight). The strain was estimated from the cross-head displacement. This method proved reasonable for measurement of uniform elongation, as was confirmed by image correlation with a CCD high-speed camera. The yield strength and ductility (taken as uniform elongation) were used as the measures in characterizing the mechanical behavior.

5.2.4 Cast and Sintered Materials

Similar procedures were followed for characterization of the cast and sintered materials. Optical microcopy and hardness testing were used for characterization.

For cast Mg AZ31 exactly the same procedures were followed with wrought AZ31. For both cast brass 360 alloy and pure sintered copper, samples were prepared with the same polishing technique and etched in a solution consisting of water, ammonium hydroxide and hydrogen peroxide (in 30% solution) in 5:5:1 volume fractions, respectively.

Vickers indentation at 100 g indentation load (Leco Hardness Tester, model: LM247AT) was used to measure hardness (strength). At least 25 indents were made in a specimen to obtain mean values and standard deviations.

The strip cross-section that is perpendicular to the chip flow direction was again the choice for orientation for optical microscopy and hardness testing. Archimedes technique was used for density measurements.

Sintered copper was donated by Powder-Tech Associates, Inc., North Andover, MA, in memory of Eugene Andreotti, who retired from Powder-Tech after 50 years of excellent service in powder metallurgy industry. It was provided in disk forms, which were cold compacted at 207 MPa, sintered at 1050 °C and then repressed at 620 MPa to the 95% relative density.

Cast magnesium AZ31 was supplied by Xi'an Yuechen Metal Products Co., Ltd., Shaanxi, China. Cast cylindrical billets were prepared by direct chill (DC) casting method. The billets were surface machined to the final diameter of 200 mm and the length was 500 mm.

Cast brass 360 was supplied by National Bronze & Metals, Inc., Houston, TX. Round billets were 125 mm in diameter and length.

CHAPTER 6. APPLICATION OF EM TO ALLOYS WITH LIMITED WORKABILITY – MAGNESIUM AZ31

6.1 General Capabilities of EM: Suppression of Segmentation, Microstructure Refinement and Influence on Mechanical Properties

Application of the EM process to MgAZ31B is most dramatically depicted in Fig. 6.1. Figure 6.1 illustrates the capability of the EM to produce sheet of high integrity and suppress segmentation in chip formation, through appropriate combinations of T and $p/2k$. While needle-like chips typical of conventional machining of MgAZ31B were produced in LP EM (Fig. 6.1a), long strips of 'defect-free' sheet were consistently created in the HP EM (Fig. 6.1b). The deformation conditions were $T = 230\text{ }^{\circ}\text{C}$ in HP EM as against $T = 70\text{ }^{\circ}\text{C}$ in LP EM, with $T_0 = 25\text{ }^{\circ}\text{C}$ (no pre-heating) in both cases. By imposing higher values of p and T in the deformation zone, the latter by using the in situ heating from the plastic work, HP EM is able to effect chip formation as continuous sheet without pre-heating of the workpiece.



Figure 6-1 MgAZ31B processed by EM: a) LP EM ($V_o = 0.2$ m/s, $p/2k = 1.1$, $T = 70$ °C) resulting in discrete needle-like chips and b) HP EM ($V_o = 1.2$ m/s, $p/2k = 1.7$, $T = 230$ °C) with chip formed as long continuous sheet. $\gamma \sim 2$, $T_0 = 25$ °C (no pre-heating), $t_0 = 0.25$ mm (after [11], used with permission of Elsevier).

After getting continuous forms of magnesium with HP EM as opposed to powder-like forms in conventional machining (peeling) or LP EM, the remaining challenge was to increase the dimensions of the strips to some commercially valuable size. While the thickness of sheet created by HP EM could likely be increased to ~ 0.5 mm without pre-heating, for example, by increasing $p/2k$ to be in excess of 1.7, such deformation conditions were difficult to access in the study due to the relatively small power capacity of the lathe (6 kW – 10 hp). However, by pre-heating the workpiece, a sheet of 2 mm thickness and 25 mm width could be created (Fig. 6.2a), in a single step, even within the power constraints. The requisite deformation conditions were achieved by setting $T_0 = 250$ °C and $p/2k = 1.7$ ($\lambda = 0.7$). Since the force (F_c) per unit cross-sectional area of sheet is essentially constant, sheet of large width can also be created within the power capacity of this EM system, provided the thickness is suitably reduced. Figure 6.2b shows an example of coiled 50 mm wide sheet of ~ 0.18 mm thickness produced in the current system. The sheet dimensions realizable in an EM system can be estimated based on the force per unit area measured for specific materials. A sheet of magnesium with 300 mm width and 1mm thickness can be produced at 1m/s with a 130 kW capacity lathe. The power

requirement is directly proportional to the deformation rate, thus same sheet at 0.5 m/s production speed will require 75 kW capacity lathe. Edge cracking in the sheet is negligible (Fig. 6.2) suggesting high yield for EM in strip manufacture.

This observation is of value in assessing scale-up of EM for it suggests that as long as segmentation is suppressed by the correct choice of process parameters, the upper limit on sheet width is determined only by EM power capacity and not by any fundamental process limitation assuming tooling stability, machine rigidity, dimensional stability, etc.

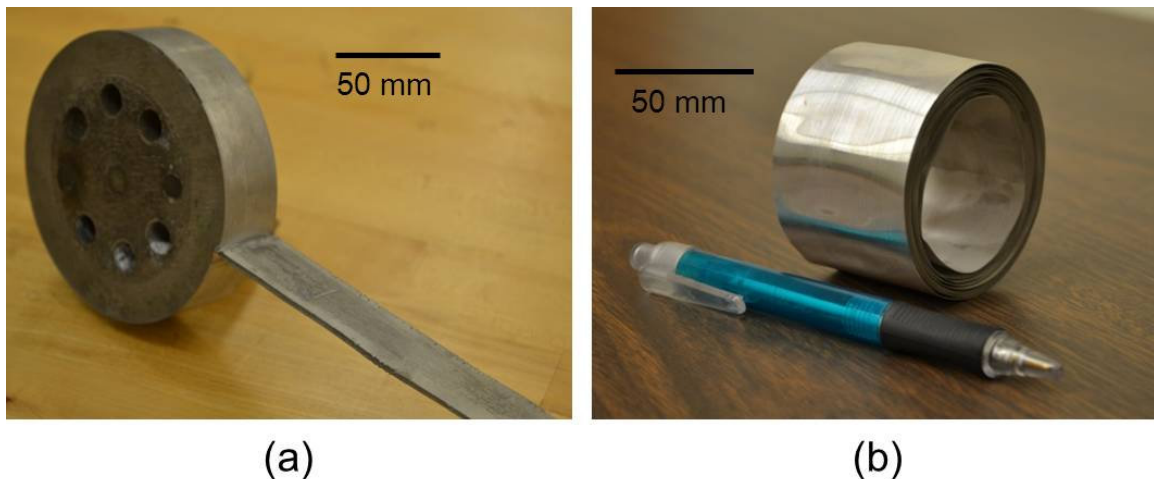


Figure 6-2 MgAZ31 sheet of large dimensions: (a) 2 mm thick x 25 mm wide sheet produced by HP EM; one end of the sheet is still attached to the bulk workpiece, (b) 0.18 mm thick x 50 mm wide coiled strip generated by LP EM. $T_0 = 250\text{ }^{\circ}\text{C}$ (after [11], used with permission of Elsevier).

Figures 6.3a shows an optical micrograph of typical sheet microstructure created by HP EM. The grain size is $\sim 3\text{ }\mu\text{m}$, demonstrating a 5-fold refinement of the grain size from that of the initial material ($16\text{ }\mu\text{m}$). The relatively uniform, equiaxed microstructure is indicative of dynamic recrystallization occurring in the deformation zone. This is likely due to increased basal slip, with some non-basal slip and very limited twinning [10, 77, 90], since $T \sim 200\text{ to }300\text{ }^{\circ}\text{C}$ and strains are large at these process conditions [50]. Similar equiaxed microstructures (grain size of 3-7

μm) were observed at other HP EM conditions resulting in sheet, with the larger grain sizes occurring at higher T . In contrast, the LP EM microstructure ($T \cong 140^\circ\text{C}$) (Fig. 6.3b) is generally not recrystallized. This microstructure consists mainly of original bulk grains, essentially preserved but highly twinned, complemented by regions of localized deformation near the sides and the edges of the chip segments. The latter regions show heavy flow lines (basal slip), but with poorly resolved microstructure likely due to cell (subgrain) formation. Thus twinning appears to dominate the deformation in the bulk of the chip, with only limited basal slip at the lower T and p [77, 90], in LP EM.

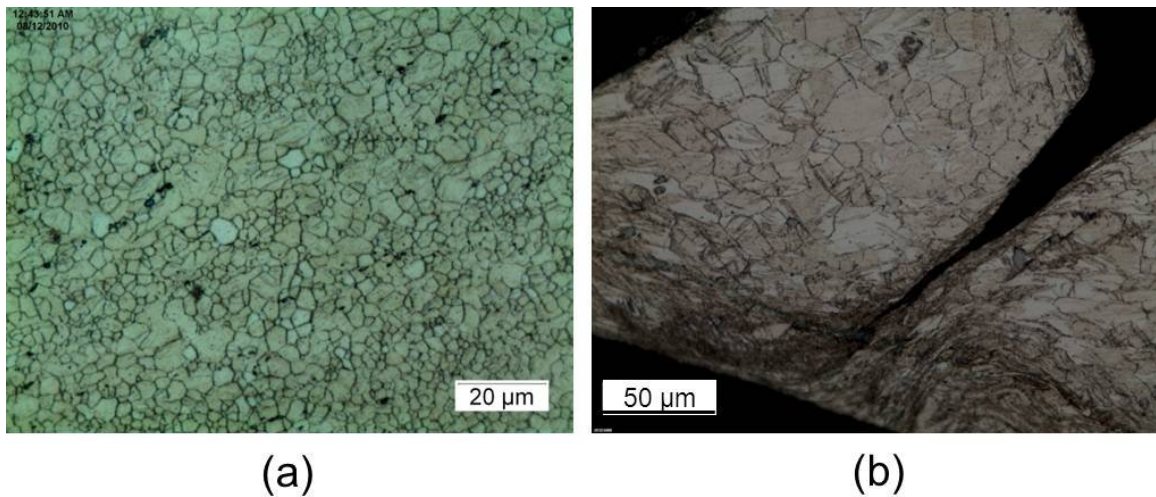


Figure 6-3 Representative optical micrographs of typical sheet microstructures: (a) HP EM (continuous) sheet produced at $T \cong 250^\circ\text{C}$. Grain size = $3.1 \pm 0.5 \mu\text{m}$. The microstructure is dynamically recrystallized and uniform across the thickness of the sheet. (b) LP EM (segmented) strip produced at $T \cong 140^\circ\text{C}$. Grains of bulk material are preserved, though highly twinned after the deformation. Note that the strain is localized around sides and edges of the segments. $T_0 = 25^\circ\text{C}$ in both cases (after [11], used with permission of Elsevier).

The Vickers hardness of the HP EM sheet of Fig. 6.3a was $68 \pm 2 \text{ kg/mm}^2$, similar to that of the initial bulk material and consistent with strain-free recrystallized grains. The uniaxial yield stress was $\sim 85\%$ that of the bulk. The measured uniform elongation was at least 40 percent greater than that of the bulk material, indicating improved ductility.

Significant additional refinement of microstructure into the UFG range (100 nm – 1 μm) could be achieved via dynamic recrystallization by lowering T and/or increasing ϵ . Figure 6.4 shows UFG microstructures in sheet (t_0 (initial chip thickness) = 0.127 mm) created by HP EM at smaller T and without pre-heating. In Fig. 6.4a, the grain size is $\sim 1.5 \mu\text{m}$ ($T \cong 205^\circ\text{C}$) while in Fig. 6.4b the grain size is $\sim 200 \text{ nm}$ ($T \cong 165^\circ\text{C}$). The latter microstructure had a hardness of $109 \pm 3 \text{ kg/mm}^2$, which is nearly twice that of the bulk material. In contrast, the somewhat coarser UFG microstructure (Fig. 6.4a) had a hardness of $79 \pm 2 \text{ kg/mm}^2$, only slightly greater than that of the bulk material. The sheet with 200 nm grain size (Fig. 6.4b) is an example of a UFG microstructure for Mg alloys, given that such microstructures are difficult to create in these alloys and, to date, have been achieved only by use of multi-step processing routes [61]. Similar grain sizes have also been observed in thicker HP EM sheet created at $T \sim 150^\circ\text{C}$, by lowering V_0 .

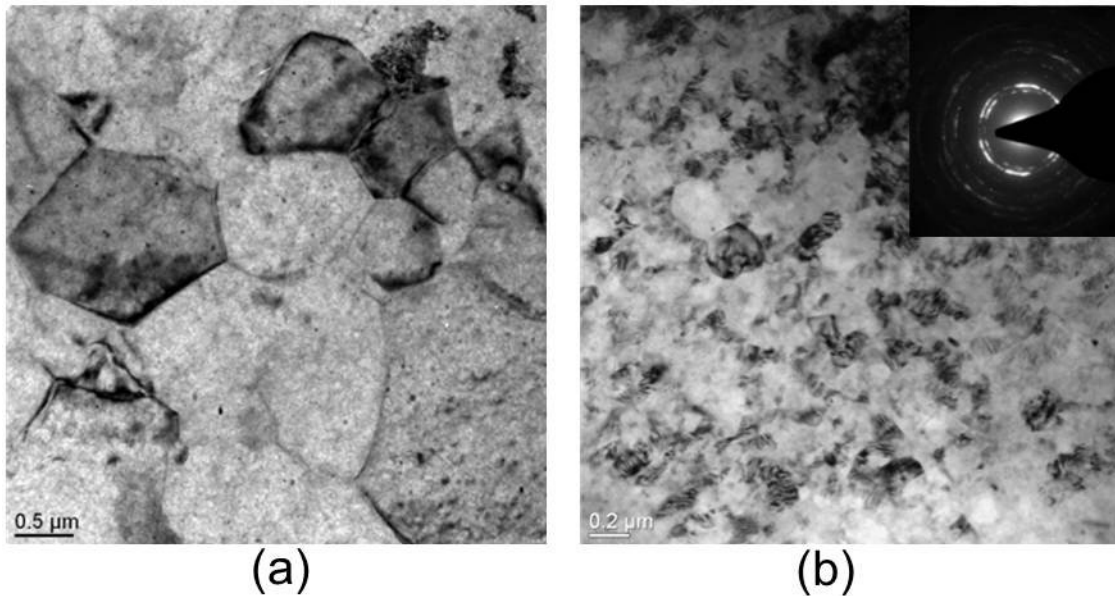


Figure 6-4 TEM bright-field micrographs of Mg AZ31 alloy showing UFG microstructures produced by HP EM at small deformation rates (velocities) (V_0): a) grain size $\sim 1.5 \mu\text{m}$ ($t_0 = 0.127 \text{ mm}$, $T = 206^\circ\text{C}$) and b) grain size $\sim 200 \text{ nm}$ ($t_0 = 0.127 \text{ mm}$, $T = 167^\circ\text{C}$). $T_0 = 25^\circ\text{C}$ for both cases. Since specific energy varies with t_0 , for $t_0 < 0.25 \text{ mm}$, the measured U_{sr} value of $5.4 \times 10^8 \text{ J/m}^3$ was used to estimate T (after [11], used with permission of Elsevier).

Besides suppressing segmentation, the interactive effects of p and T are of value in controlling sheet microstructure. Grain refinement from an initial value of $16\text{ }\mu\text{m}$ in the bulk alloy to $\sim 3\text{ }\mu\text{m}$ in the sheet has been consistently realized using HP EM with no pre-heating (Fig. 6.3a). The microstructure shows highly equiaxed grains, consistent with the dynamic recrystallization to be expected at these deformation conditions ($T \sim 230\text{-}280\text{ }^{\circ}\text{C}$). Basal slip together with non-basal slip appears to dominate the deformation at these conditions with only limited twinning. The occurrence of this slip and dynamic recrystallization is supported also by recent texture measurements, and prior studies [10, 50, 77, 90]. When T in HP EM is lowered to $150\text{-}200\text{ }^{\circ}\text{C}$ by reducing V_0 , continuous sheet forms with UFG ($\sim 200\text{ nm}$ to $1.5\text{ }\mu\text{m}$) microstructures as shown in Figs. 6.4a and 6.4b. The 200 nm UFG microstructure is found to have a high hardness of $\sim 110\text{ kg/mm}^2$, almost twice that of the initial bulk material. Hydrostatic pressure again plays a key role by facilitating workability and suppressing segmentation, even when T is only $150\text{ }^{\circ}\text{C}$ and below the oft quoted temperature of $\sim 200\text{ }^{\circ}\text{C}$ for enhanced slip system activity in this alloy. It should be noted that the time for which the material is exposed to T is quite short $\sim 0.1\text{ ms}$. The short time exposure is favorable for minimizing grain growth; however its implications for phenomena such as recrystallization and recovery need to be explored. Other characteristics of HP EM sheet include a relatively uniform microstructure across the sheet thickness at all of the process conditions, likely due to the uniform deformation field prevailing in EM (Fig. 2.2); and edge cracking levels ($\sim 5\%$ of sheet width) smaller than in rolled sheet. These are of value for production applications.

In the following sections, the effect of process parameters on the sheet dimensions, quality, microstructures and the micromechanics of formation of the UFG will be detailed.

6.1.1 Effects of Hydrostatic Pressure and Deformation Zone Temperature on Segmentation

At a given initial chip thickness ($t_0 = 0.25$ mm), T can be adjusted independently of $p/2k$ by appropriate selection of V_0 and T_0 , enabling a separation of their effects on workability. Figure 6.5 shows micrographs of longitudinal (and through sheet thickness) cross-sections of sheet made at different levels of p and T . With LP EM, even a T as high as 280 °C is not sufficient to eliminate serration present in the portion of the chip that is in contact with the constraining edge. At this point, it is important to make the distinction between serration and segmentation. Serrations in this thesis are defined as minor cracks that occurs only on the top surface of the strips, not penetrating deeper to the strip thickness. They are often observed in LP EM, where constraining tool barely touches the top surface. These crack are not detrimental to the strip quality, although they need to be removed before the further processing of the strip. Segmentations, on the other hand, are defined as large cracks usually begin from the top surface and penetrate at least he half thickness of the strip. Segmentations, both possible in HP and LP EM due to insufficient plasticity and shear banding, usually result in disintegration of the strips.

Such segmentation/serration has been observed in machining of Mg alloys [91], and even in ECAP, under many conditions where the workpiece was pre-heated [45, 63]. However, with HP EM, even $T = 230$ °C ($T_0 = 25$ °C) is sufficient for producing defect-free sheet. This illustrates the important role of p in enhancing workability and suppressing segmentation and serration. Furthermore, none of the HP EM conditions shown in Fig. 6.5 involved pre-heating of the workpiece; the requisite deformation temperatures were achieved purely by utilization of the plastic dissipation. In contrast, realization of $T > 135$ °C in LP EM required pre-heating.

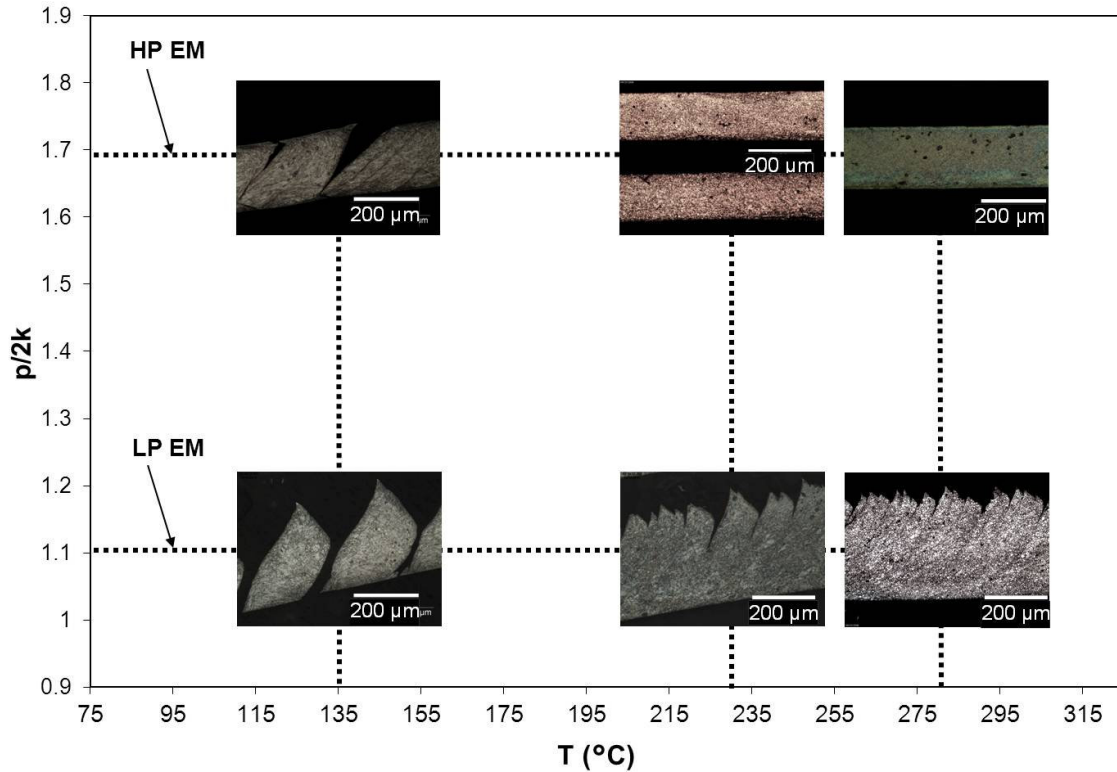


Figure 6-5 Longitudinal cross-sections of sheet created under different conditions of $p/2k$ and T . The effect of hydrostatic pressure in enhancing workability and suppressing segmentation is seen in the sheet created by HP EM at the higher T conditions (after [11], used with permission of Elsevier).

There should be a minimum temperature, around 230 °C, established in the workpiece in order to enhance ductility and continuous strip formation. This minimum temperature can be attainable either by sufficient deformation heating through increased deformation rate and/or by initial pre-heating. When the temperature and pressure effects studied independently it has been observed that the minimum temperature desired for the sufficient plasticity decreases by the application of hydrostatic pressure.

6.1.2 Effect of Initial Chip Thickness on Segmentation

Chips created by conventional machining tend to segment when t_0 is increased. This segmentation is typically due to fracture or shear banding [88, 92]. Commonly used

approaches to suppress segmentation in machining and ECAP are by pre-heating the workpiece/tool, and/or application of back-pressure [45, 63, 90]. Instances of segmentation of Mg AZ31B chips have already been seen in LP EM. Since this phenomenon can be a limiting factor in the deformation processing, the effect of increasing t_0 on sheet formation was studied. A constant $V_0 = 1$ m/s and $T_0 = 25$ °C were used with HP and LP EM. This ensured that T was approximately constant over the range of t_0 within each of the process configurations ($T \cong 150$ °C for LP EM and $T \cong 250$ °C for HP EM). Figure 6.6 shows macrographs of chips created at select t_0 . The transition from a continuous sheet to a segmented chip is seen to occur at a much larger t_0 (~ 0.5 mm) in HP EM than in LP EM ($t_0 \sim 0.1$ mm). Even though larger sheet thickness without segmentation can be realized in LP EM by making T equal to that in HP EM, minor serration is still unavoidable. This once again demonstrates the role of increased p (HP EM) in suppressing segmentation and serration, thereby, enabling larger thickness of sheet to be realized. The use of hydrostatic pressure, rather than workpiece pre-heating, to extend the process limits for continuous sheet production is likely to have important benefits also for creating UFG sheet forms by EM.

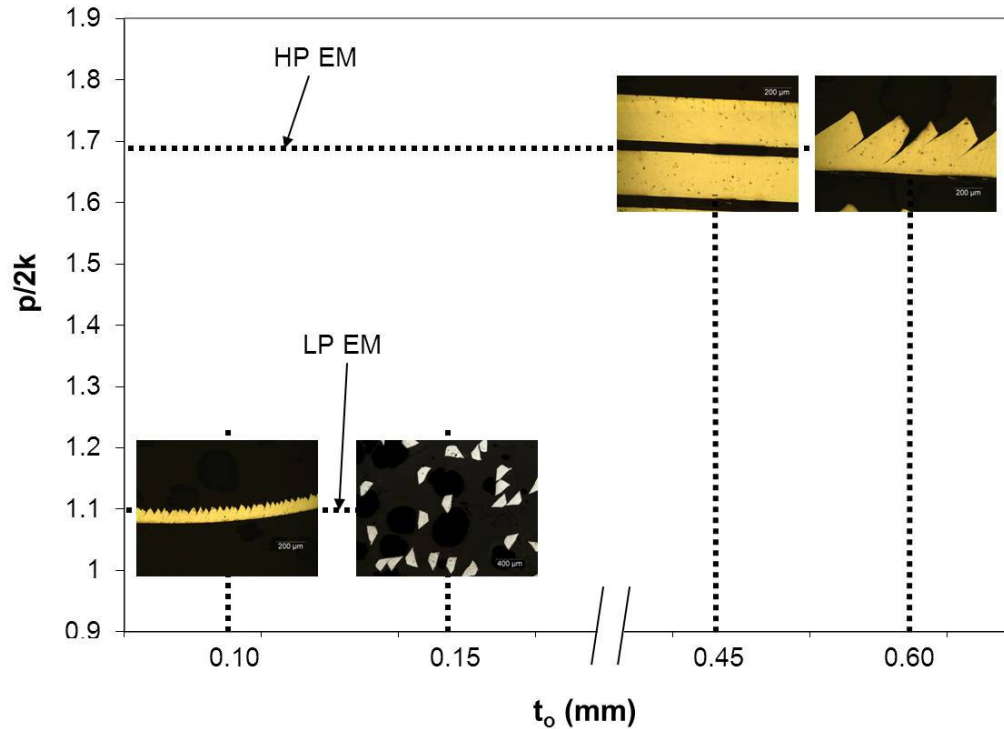


Figure 6-6 The effect of hydrostatic pressure on chip segmentation at different t_o . $V_o = 1$ m/s and $T_0 = 25$ °C. Note that a continuous sheet of larger thickness is realized by suppressing segmentation in HP EM than in LP EM (after [11], used with permission of Elsevier).

6.1.3 Guidelines for Continuous Strip Formation with Uniform Microstructures

The various examples of deformation processing show that high p levels in conjunction with control of T , using the in situ plastic dissipation, can be effectively used to produce sheet, foil and strip of commercial thickness in a single step of deformation. Some guidelines for practical use of EM in deformation processing may now be suggested.

There appears to be two type of segmentation during chip (strip) formation:

- 1) Segmentation and serration at constant t_o due to the insufficient temperature and/or pressure in the deformation zone. Flow (shear) localization and crack propagation events are associated with segmentation. Sheets have very poor quality and sometimes discrete segments are obtained instead of continuous

forms. Whereas, the serration event is less detrimental for sheet quality (Fig. 6.5). Flow localization is not always associated with the serration.

- 2) Segmentation at constant temperature and pressure due to increased t_0 . Flow softening is not necessarily the cause of this type of segmentation. Once the critical t_0 is achieved and passed continuous forms are not realizable, discrete segments are observed (Fig. 6.6).

For the first type of segmentation, flow localization is caused by adiabatic shear band formation due to the interactive effects of temperature and strain. Periodical shear bands that are forming across the thickness of the sheet, separate the strain localized regions from the relatively strain free matrix (Fig. 6.3b). This behavior is observed both in HP and LP EM. Cracks are usually formed across shear bands and they are the reason of discrete particle formation and disintegration of the sheet. Serrations, on the other hands, form only in LP EM. Crack formation is rare in serrated strips and continuous forms can be obtained in spite of the fact that top surface finish would be detrimental for most practical applications. The intensity of the localization seems to be decreasing with increasing temperature for both HP and LP EM (Fig. 6.5). Therefore, the deformation mechanisms operating at different temperatures and/or strain rates may play a key role in shear banding behavior.

Figures 6.3b and 6.7 compare the segmentation and serration behavior. Typical segmentation with adiabatic shear bands is demonstrated for LP and HP EM in Figs. 6.3b and 6.7a, respectively. Original bulk grains are preserved in the matrix in both cases and strain (shear) is localized to the edges and the sides of the segments. Deformation temperature is lower than 200 °C and around 140 °C in both cases, and the less strained bulk grains appear to be twinned and the localized regions are dynamically recrystallized to the ultrafine regime. The strain was targeted to be $\varepsilon = 1.15$ for both cases, however the HP EM strip seems to be more strained and refined. The reason for the strain differences can be that the natural chip thickness

ratio was not realized. In LP EM, the chip thickness ratio was targeted to be 1.4 which is the natural maximum ratio attainable in magnesium and results in the effective strain of 1.15. However, this natural chip thickness ratio is dependent on deformation temperature and the value of 1.4 is only realizable at high temperatures ($> 200\text{ }^{\circ}\text{C}$). The actual ratio in fig. 6.3a is only 1.1 which results in a strain of 1.06. Moreover, application of hydrostatic pressure in HP EM may result in activation of additional slip systems at the same temperature and this can result in more refinement of the microstructure [40].

The serration behavior is shown in Figure 6.7b. The microstructure is more uniform compared to the shear banding and distinct dynamic recrystallization is observed with fine grain size. Grains are equiaxed in shape and appear to be strain-free in the optical microscopy. Cracks are not observed and shear localization is minimal, although sometimes the grain size along the serrations is finer than the matrix. This may be an indication of weak localization.

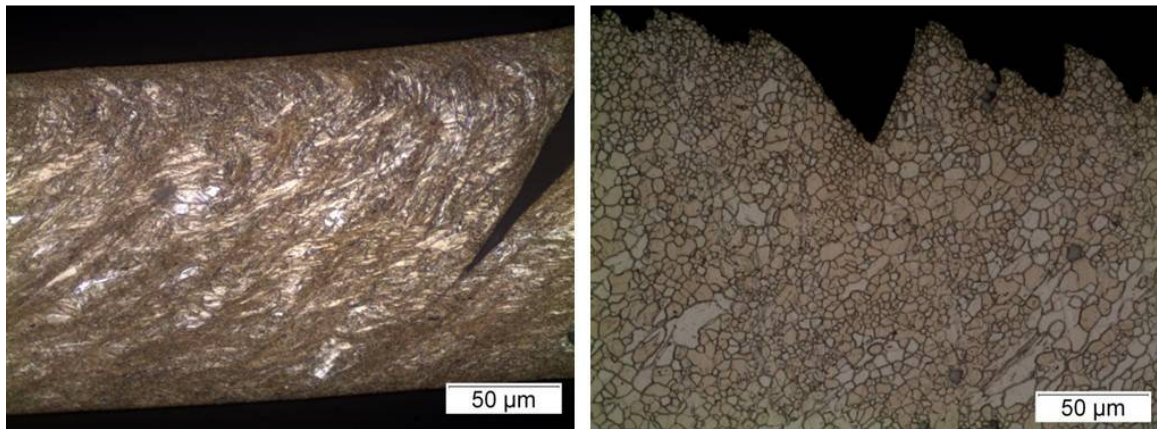


Figure 6-7 Representative optical micrographs of strips with defects: (a) HP EM (segmented) sheet produced at $T \cong 140\text{ }^{\circ}\text{C}$. The microstructure is shear banded and cracks grow through shear bands. (b) LP EM (serrated) strip produced at $T \cong 260\text{ }^{\circ}\text{C}$. Dynamic recrystallization is evident with somewhat finer grains through serrations. No crack formation and growth along the thickness.

The second type of segmentation occurs with an increase in initial chip thickness (feed) (Fig. 6.6). According to the temperature rise models, the temperature in the deformation zone is little affected by the increase in feed. The heat partition coefficient increases with feed, causing more temperature rise. Therefore, according to the load stability models, adiabatic shear banding cannot be the reason of segmentation with increasing initial chip thickness.

In this thesis, this type of segmentation will not be investigated any further but several experiments may be suggested to test the potential causes. The width of the deformation zone (Δ) may change with t_0 . Then, the strain rate and the temperature in the deformation zone would be affected and adiabatic shear banding would become a possible mechanism. Hydrostatic pressure may also decrease with an increase in Δ , which may ease crack formation. An ideal experiment for the future work would be PIV measurements of the shear zone thickness with a variation in t_0 . Temperature and strain rate can be remodeled with the measured values and shear banding possibility can be tested. As it mentioned in Section 3.3, perturbation analyses models can incorporate the thickness effects of the shear zone [45]. As the shear zone thickness increases, the number of possible wavelengths in this finite zone will increase. Now, the perturbations may have enough time to grow to the maximum extent which is described as shear banding. The critical time for some perturbations growing to the maximum extent will decrease, increasing the susceptibility to shear banding. Therefore, measurement of Δ with t_0 will be of crucial for testing the possibility adiabatic shear banding.

Besides PIV, microstructural techniques can also show the thickness of the zone with t_0 . A simple experiment is the investigation of the head section of the strips. When extrusion machining tool first engages with the workpiece the feed will slowly increase and reach to the preset value. The head section of the strip can be mounted and examined with optical microscopy and the microstructures can be investigated

as a function of head in order to understand the effect of microstructural mechanism (if any) on the segmentation.

Alternatively, the thickness can be measured by “quick-stop” experiments. The process can be interrupted after reaching to steady-state (i.e. when the feed is saturated) and the strip that is still attached to the bulk can be examined. The thickness of the deformation zone can be measured by comparing the initial microstructure to the final strip microstructure. The transient region between those two regions will be the deformation zone.

6.2 Effects of Strain Rate and Temperature on Shear Band Formation and Dynamic Recrystallization

In this section, the effects of high strain rates and low temperatures on the shear banding behavior and microstructure evolution are investigated. This thermo-mechanical domain is selected due to its relevancy to the cost and energy of manufacturing strip with the possibility of the undesired phenomenon of shear banding. Section 4.2 serves as an introduction and background for this section, as will be referred in the discussions.

Processing maps are developed and optimal thermo-mechanical conditions are established for creating sheet with uniform microstructures and devoid of shear bands. Grain size is controlled in the ultrafine and fine grain regime, based on understanding of deformation mechanisms and dynamic recrystallization operative in these thermo-mechanical domains.

6.2.1 Controlling Strain Rate and Temperature in the Deformation Zone:

Experimental Framework

Effects of hydrostatic pressure and initial chip thickness are established in the previous sections. It is clear that decreasing chip ratio and increasing hydrostatic pressure drastically increase the strip quality in magnesium. Therefore, a constant λ

$= 0.7 \pm 0.1$ (HP EM) and a constant initial chip thickness of 0.250 mm are selected to study the interactive effects of temperature and strain rate. The constant λ also fixes the equivalent strain to 1.13 ± 0.07 at a fixed rake angle of 5° .

There were two types of experiments. In the first experiment, the deformation rate was varied at room temperature from 0.15 to 4 m/s; which both changed the strain rate and temperature in the deformation zone simultaneously (Figure 6.8top). Temperature rises in the deformation zone with increasing deformation rate (Figure 5.2). Blue line in Figure 6.8 shows this relationship between the deformation rate and temperature for $T_0 = 25^\circ\text{C}$. The cutting force at each deformation rate is measured and the temperature is calculated using Eqs. 52 and 53. Red squares show the calculated effective strain rate ($d\bar{\epsilon}/dt \cong \bar{\epsilon}V/\Delta$) at each corresponding deformation rate.

In the second experiment, the combined effects of the deformation rate and temperature were separated by selecting constant deformation rates and varying the initial temperature (T_0) between 25 and 250 $^\circ\text{C}$ at each fixed deformation rate. There were three constant strain rate experiments at 0.15, 0.5 and 1 m/s. Assuming the deformation zone thickness stays constant at 55 μm [22, 44] with the deformation rate, equivalent strain rates of 3×10^3 , 1×10^4 and 2×10^4 were realized. It was not possible to separate the strain rate and temperature effects in the first experiment since they both increased simultaneously due to the plastic heating and the temperature rise in the deformation zone. In the second experiment, it was possible to study the temperature effects solely, by fixing the strain rate and increasing the pre-heating temperature (T_0) from the ambient. Alternatively, the deformation temperature (T) can be also fixed at $\sim 230, 280$ and 320°C and strain rate can be varied in order to study the strain rate effects (Figure 6.8bottom). The strain rate was varied by only an order of magnitude in the current work, future experiments are recommended for a wider range of strain rates.

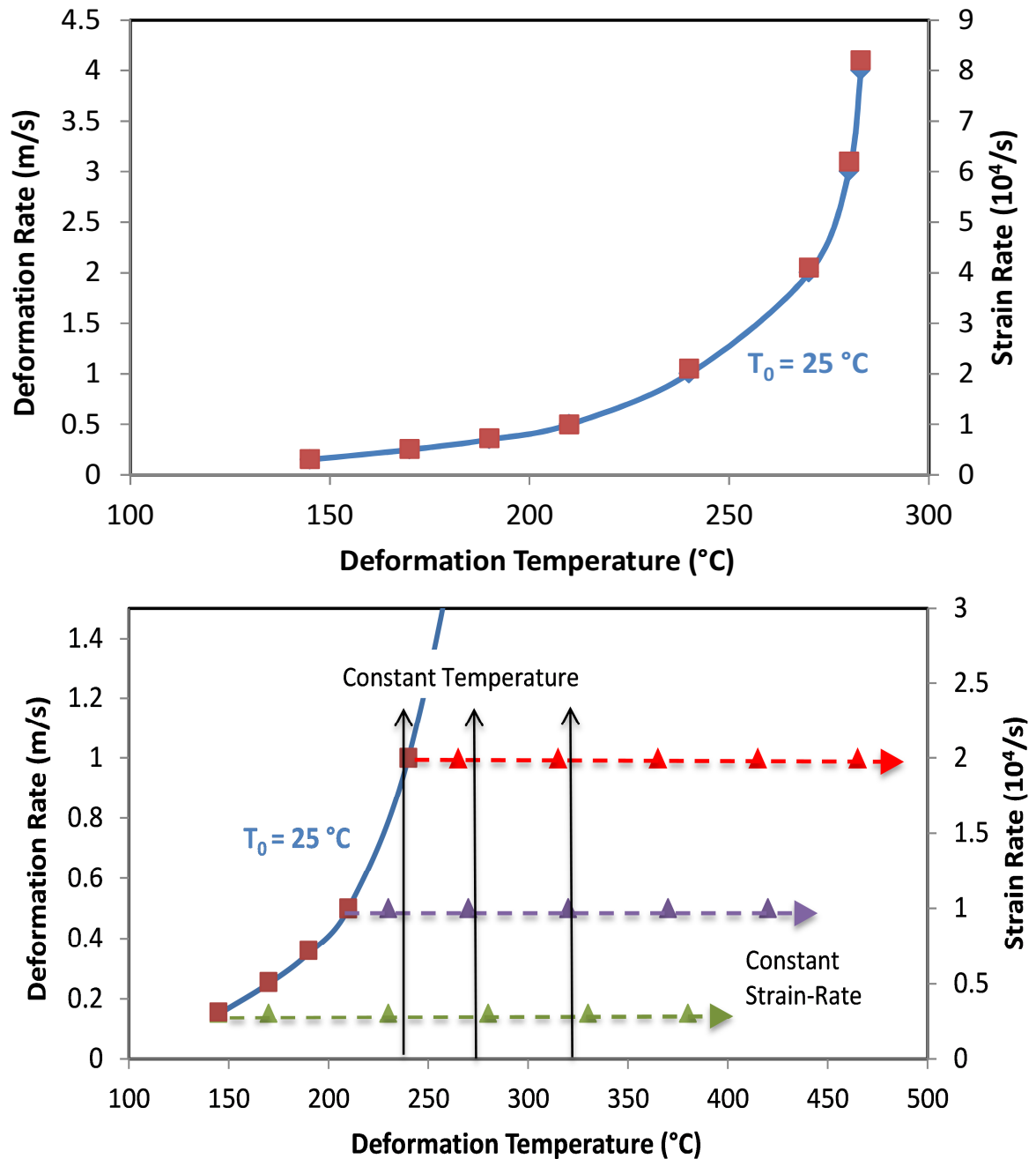


Figure 6-8 Interactive strain rate and temperature experiments. Deformation rate and temperature are varied (increased) simultaneously at $T_0 = 25^\circ\text{C}$ in the first experiment (top). Blue line shows the temperature increase with the deformation rate. Red squares represent the calculated strain rate (plotted on the second y axis) at a given deformation rate. Deformation rate kept constant and deformation temperature varied in the second experiment (bottom). It was possible to separate the effects of temperature and strain rate.

Free machining (LP EM) experiments (-where constraining tool was barely touching the top surface) were also conducted to realize lower temperatures, since the shear energy of LP EM is smaller than HP EM and the temperature rise is limited. The effective strain was kept constant at 1.15, same as HP EM. As discussed in Section 6.1, the hydrostatic pressure in LP EM ($p/2k = 1.1$) was lower than HP EM ($p/2k = 1.7$). Again the strain rate was fixed to $5.2 \times 10^3/s$ (0.25 m/s) and the pre-heating temperature was increased. Thus, the deformation temperatures of 80, 105, 155, 205 and 255 °C were possible (not shown in Fig. 6.8). Results from this experiment are not intended to be directly compared to HP EM, but they will give insights on low temperature deformation behavior of magnesium and they will establish guidelines for the future experiments on low strain rate ($10^2/s$) - low temperature (< 150 °C) EM.

6.2.2 Optical Microstructures and Shear Banding Behavior

Figure 6.9 summarizes the shear banding behavior of the HP EM strips with increasing deformation rate. As deformation rate increases, the deformation temperature simultaneously increases, resulting in continuous strips. There appears to be a critical temperature (shown by the dotted vertical line) above which shear banding is not observed. At low temperatures, cracks are usually associated with shear bands. Microstructure is highly inhomogeneous with strain localized to shear bands (unresolvable regions in optical microscopy) and less strained matrix where original bulk grains are preserved. Although not resolvable under optical microscope, the shear bands are later shown to be dynamically recrystallized (DRX) by TEM. When the critical temperature is passed, the DRX event is observed throughout the microstructure with remnants of the original grains. The DRX grains get bigger, remain equiaxed and strain-free with further temperature increase. At higher strain rates ($8 \times 10^4/s$ – not shown in figure 6.9), very similar microstructures are obtained at a slight increase in the DRX grain size with the temperature rise. Based on these preliminary observations, the shear banding phenomena seems more sensitive to the deformation temperature rather than the

strain rate. Even at the maximum strain rate, $\sim 10^5/\text{s}$, the shear banding is not visible and the microstructure is fairly homogeneous.

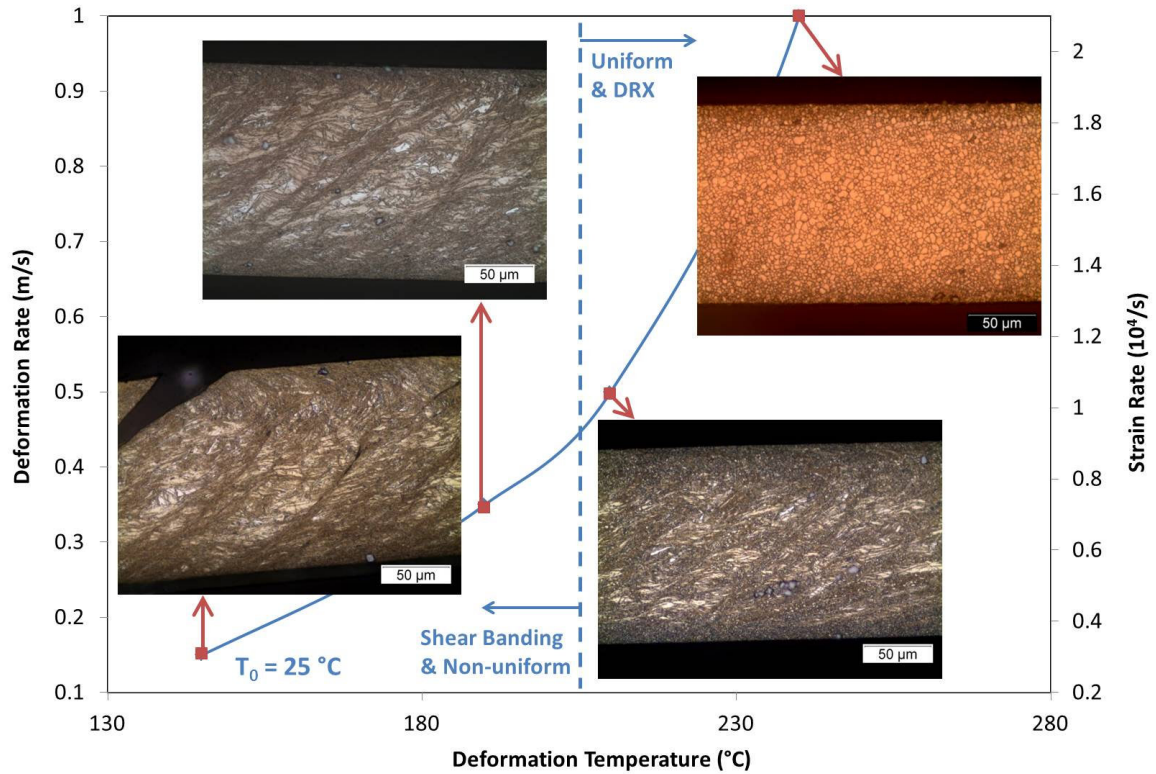


Figure 6-9 Shear banding behavior in HP EM ($\epsilon = 1.2$) with increasing strain rate and temperature. Blue line shows the temperature increase with the deformation rate at $T_0 = 25\text{ }^\circ\text{C}$. Red squares represent the calculated strain rate (plotted on the second y axis) at a given deformation rate. Representative optical microstructures of the strip cross-sections are given for each strain rate and corresponding temperature. Note that the deformation rate and the strain rate axes are drawn in scale to each other.

In order to separate the strain rate effects and to isolate the temperature effects on the shear banding, a processing map is developed based on microstructure observations for constant strain rates and temperatures (Fig. 6.10). It is clear that, even at a constant strain rate, the shear banding is highly dependent on the temperature. The critical temperature again appears to be $\sim 230\text{ }^\circ\text{C}$ in the case of HP EM. The critical temperature depends on the strain and hydrostatic pressure. Based on our observations, the critical temperature increases with decreasing hydrostatic

pressure at a constant strain (Figures 6.5 and 6.7b). The strain dependence of the critical temperature will be explained later in Section 6.2.4 based on the findings from TEM studies.

Further increase in the temperature, increases the DRXed grain size with more strain relaxation. The grain size also gets bigger with an increase in strain rate at constant deformation temperature. This difference in the grain size can be explained by local deviations between the targeted and actual strain and hydrostatic pressure values due to the experimental variations in setting up the chip thickness ratio in the extrusion machining tool. Irrespective of the deviations, a very fine DRX grain size on the order of 1-3 μm can be obtained once the sufficient deformation temperature is realized. This grain size appears to be stable within 100 °C above the critical temperature.

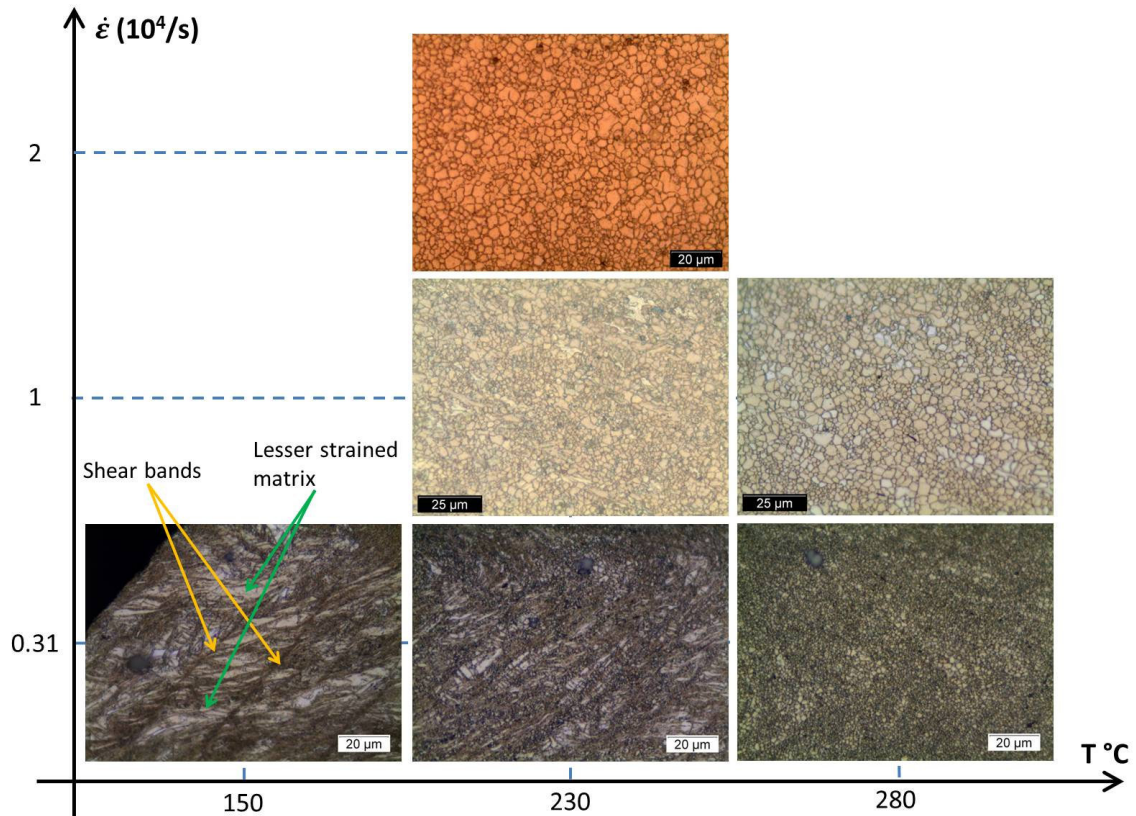


Figure 6-10 Processing conditions-microstructure map ($\epsilon = 1.15 \pm 0.07$) for HP EM at various strain rates and temperatures. The processing conditions correspond to the ones in Figure 6.8bottom.

In order to further understand the low temperature deformation behavior, the constant strain rate experiments are repeated for LP EM (Figure 6.11). Discrete particles are obtained at temperatures < 150 °C (Figs. 6.11a&b). Formation of these particles can be attributed to the shear banding and cracking behavior. This time the shear banding is so severe that cracks propagate along the entire cross-section causing the disintegration of the strip. Twinning appears to be dominant in the matrix where original bulk grain are deformed. Segmentation becomes less severe when temperature is increased. The original grains are still preserved but the DRX grains inside the shear bands become resolvable. Also, the fraction of the recrystallized regions is increased. Twinning is still observed in the original grains albeit to a lesser extent. This suggests that, dislocation activity increases and slip

becomes dominant with the temperature increase. It is important to note that, the transition from twinning dominated flow to dislocation and DRX dominated flow appears to be occur around 200 °C, the often-quoted temperature in magnesium. Twinning completely diminishes when temperature is above 250 °C (Fig. 6.11d), without any shear banding and segmentation. Only minor serration is observed with complete DRX resulting in well equiaxed and strain-free grains. Microstructure becomes more homogeneous with a mixture of very fine DRX structure and some larger elongated grains which are suspected to be the original grains deformed to a lesser strain.

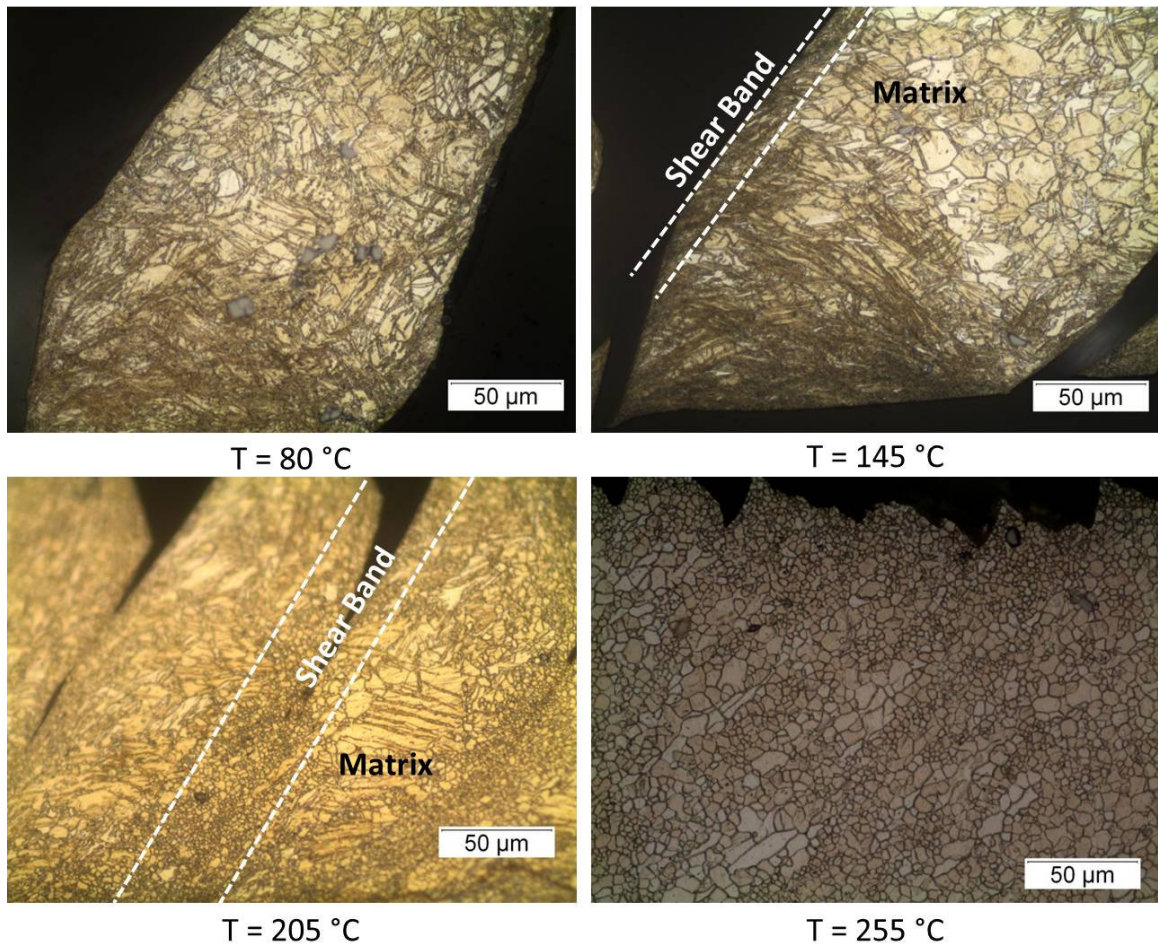


Figure 6-11 Constant strain rate and strain ($\epsilon = 1.15$) experiment for LP EM. Strain rate is fixed to $5.2 \times 10^3/s$ (0.25 m/s) and T_0 is varied to control the deformation temperature. Representative optical micrographs show the microstructure evolution and shear banding behavior with the temperature.

6.2.3 Hardness Measurements

Vickers hardness measurements from the microstructures presented in the previous sections are superimposed on the processing map in Figure 6.12. Non-uniform and shear banded microstructures have higher hardness compared to their uniform counterparts (same strain rate). At a given strain rate, hardness decreases as temperature increases and microstructure becomes fully DRX. Again, deviations

in the actual strain and hydrostatic pressure values resulted in differences in hardness at a given deformation temperature.

Hardness of the EM strip at low temperature almost doubles (99 HV) the as-received value, indicating severe plastic deformation and ultrafine grain (UFG) structure formation. Indeed, the DRX grain size and the fraction of the recrystallized regions have a significant effect on the hardness. At low temperatures, the DRX grain size is unresolvable (UFG) by optical microscopy and microstructure looks severely strained. As temperature increases, the fraction of the DRX regions increases and the recrystallized grain size increases by a factor of 5, which lowers the resultant hardness by 30%. Moreover, hardness drops rather smoothly with temperature and no sudden drop is observed when switching from non-uniform to uniform microstructures suggesting that the DRX phenomena is occurring at all the deformation temperatures with the fraction of DRX increasing continuously with the temperature.

These results indicate a smooth increase in the extent of DRX and associated softening with increasing deformation temperature. However, it is hard to conclude that the shear banding is controlled by the hardness and/or the DRX activity. The DRX appears to occur at all temperatures with the fraction increasing with temperature. There are conditions that have high hardness (93 HV) and very fine grain size of $\sim 1 \mu\text{m}$, which are lacking of shear banding (Figures 6.10 and 6.12, $\dot{\epsilon} = 0.3 \times 10^4/\text{s}$ and $T = 230 \text{ }^\circ\text{C}$). The microstructure itself can be severely strained without the formation of shear banding. Rather the shear banding seems to be controlled solely by the deformation temperature and the strain rate itself has only minor effects. Even the variations in the actual strain and hydrostatic pressure have minor effects on shear banding once the critical temperature is exceeded. This can be clearly seen in Figure 6.10 at the constant temperatures of 230 and 280 $^\circ\text{C}$. Microstructures are quite different due to the actual strain differences but they do not show any indication of shear banding.

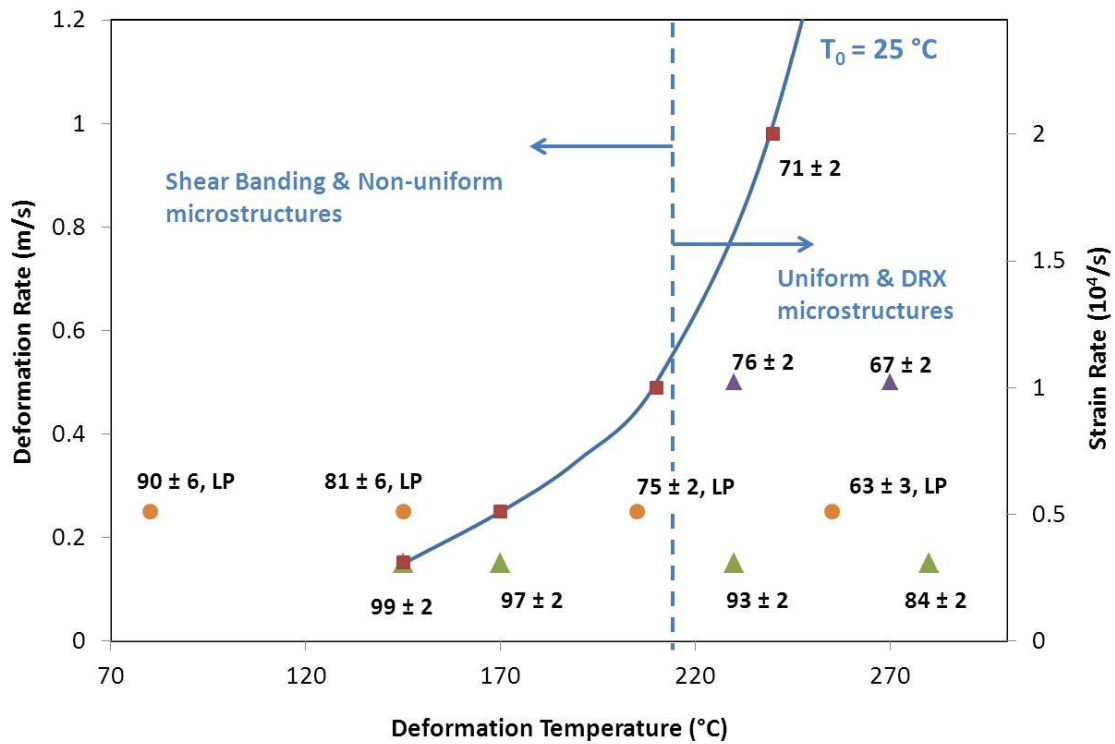


Figure 6-12 Processing map with the hardness values (Average $\pm$ Standard Deviation) measured at various strain rates and temperatures. All conditions are HP EM unless noted as LP.

6.2.4 Micro mechanisms of Deformation: Microstructure and Texture Evolution

The deformation behavior and microstructure evolution is further investigated by the help of TEM and EBSD. Figure 6.13 compares the TEM microstructures obtained at various strain rates and temperatures. For the conditions resulting in shear banding (Fig. 6.13A-F, left side), the strain rate is varied between $5 \times 10^3/\text{s}$ and $4 \times 10^4/\text{s}$. Temperature is fixed around 150°C for different strain rate experiments (Figure 6.13A-D). Then, the temperature is increased to 200°C for the fixed strain of $5 \times 10^3/\text{s}$.

The microstructure is a mixture of heavily deformed regions having high dislocation density and DRX grains for both strain rates (Figure 6.13 A&C). There is a distinct separation of these two regions in Figure 6.13C. Although DRX is reported to occur at temperatures $> 200\text{ }^{\circ}\text{C}$ in magnesium [10, 50], EM microstructures show DRX at both strain rates and low deformation temperature ($\sim 150\text{ }^{\circ}\text{C}$) with ultrafine grain size of 150 nm. DRX grains have equiaxed shape and appear to be relatively strain free (Fig. 6.13B&D).

Further increase in the temperature at constant strain rate increases the fraction of DRX regions, yet the microstructure is still a mixture of high dislocation density (non-DRX) and DRX regions (Fig. 6.13C-F). Dislocations tend to organize and rearrange into cell and subgrain walls in the heavily dislocated regions leaving the matrix almost free of dislocations (Fig. 6.13F). Once the necessary temperature ($T > 230\text{ }^{\circ}\text{C}$) is realized, microstructure becomes almost fully DRX with very fine ($0.5 - 1\text{ }\mu\text{m}$) and strain free grains (Fig. 6.13G&H , right side). Uniform flow without shear banding is possible at these temperatures. The grain morphology at these high temperatures resembles the ones in low temperature, both having equiaxed shape and being relatively strain-free. The only visible difference seems to be the size of DRX grains which increases with temperature irrespective of the strain rate (Figs. 6.13B, D&H).

Microstructure at high strain rate and high temperature (Fig. 6.13H) is fully DRX and strain appears to be annealed out completely. On the other hand, some dislocated regions are still present at the same deformation temperature, but at lower strain rates (Fig. 6.13G). The most likely reason for this difference is the variation in actual strain rather than the strain rate. As mentioned previously, the sample in Fig. 6.13G was strained to a slightly higher value due to difference between the actual and target chip thickness ratio. However, the shear banding behavior does not depend on the variations in the strain rate and/or strain once the

critical temperature is realized and the DRX becomes dominant in the microstructure.

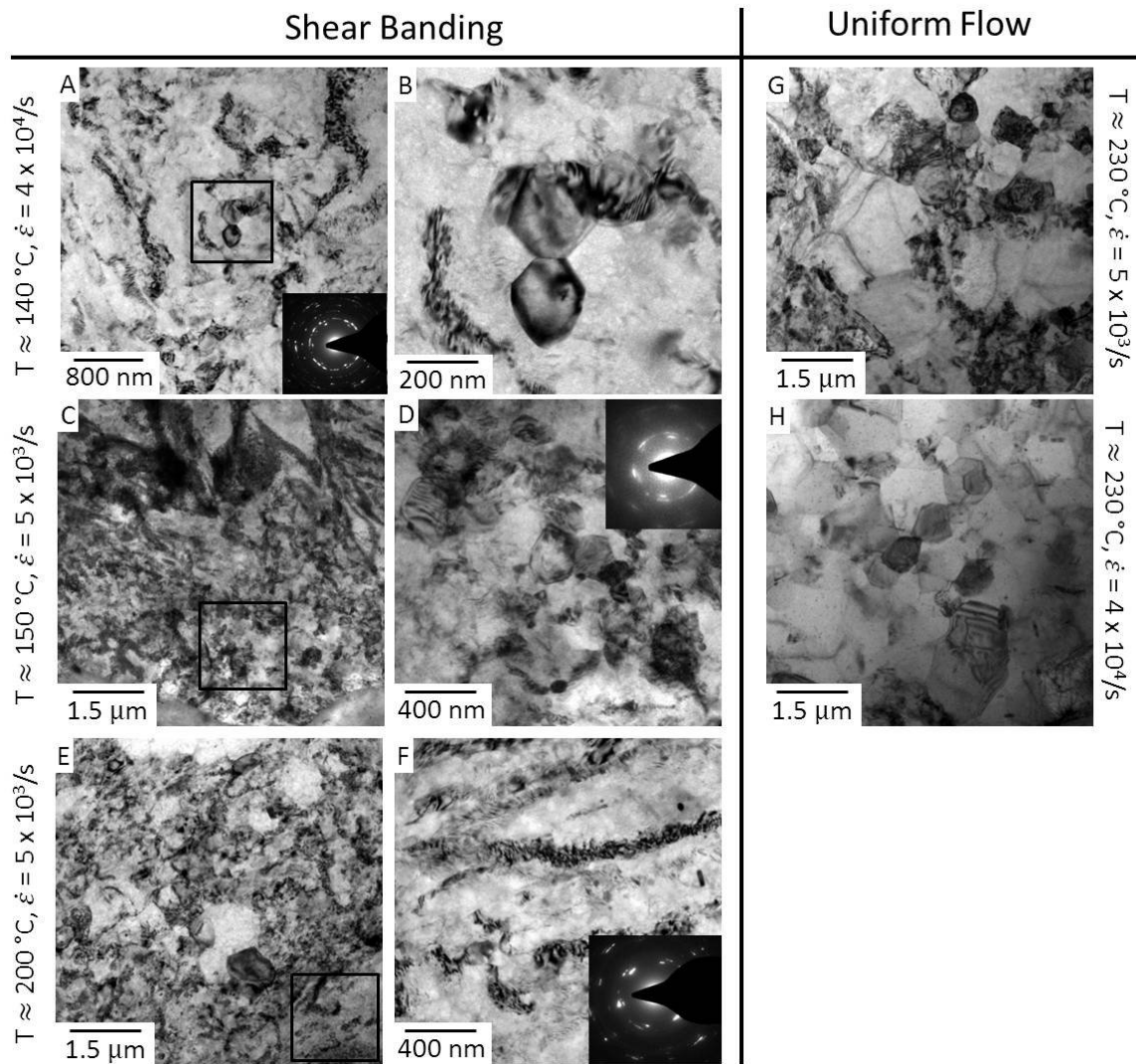


Figure 6-13 TEM bright field microstructures and diffraction patterns at various strain rates and temperatures. Diffraction patterns are taken from the entire area of the images shown. Black squares on the low magnification images (A, C and E) are showing the location of corresponding high magnification images (B, D and F). Non-uniform microstructures that are typical to shear banding (A-F) are a mixture of heavily dislocated regions and DRX regions. DRX is almost complete at higher temperatures and results in uniform flow (G-H).

In order to sample a large area in the strips, a series of images are taken over a larger area and montaged for the condition shown in Fig. 6.13E. An area of 7.5 x 97.5

μm is covered along the middle thickness of the strips that is parallel to the surfaces (i.e. electropolishing from the strip surfaces allowed to image the center of the strips). The average width of the shear bands is measured to be between 4 and 10 μm from the optical microscopy (Fig. 6.10, $T \approx 150^\circ\text{C}$). Therefore, both the shear bands and the lesser strained matrix regions can be examined.

For this experiment, the deformation temperature was $T \approx 200^\circ\text{C}$ and the shear bands were not as distinct as the low temperature conditions. The optical microstructure was still non-uniform, however, with a mixture of higher strained regions appearing as vague shear bands and lesser strained matrix between the shear bands (Fig 6.9 – second microstructure from left). For a future study, a similar large area TEM can be performed on the lower temperature strips ($T \leq 150^\circ\text{C}$), where distinct shear bands are observed together with drastically different (i.e. almost non-strained) matrix. This study may be more challenging than the current one, as the strips will be segmented and harder for sample preparation. Focused ion beam (FIB) technique can be employed to lift-off samples from the shear bands and the matrix.

Twins are not observed in any of the TEM microstructures, suggesting that the dominant deformation mechanism is the dislocation slip. Even the lesser strained matrix regions show considerable dislocation activity accompanied with some DRX (Figure 6.14). The microstructure inside the shear band is similar to Figure 6.13E. Therefore, with increasing temperature DRX starts to happen everywhere in the microstructure resulting in more uniform microstructures and weaker shear localization appearing as vague shear bands.

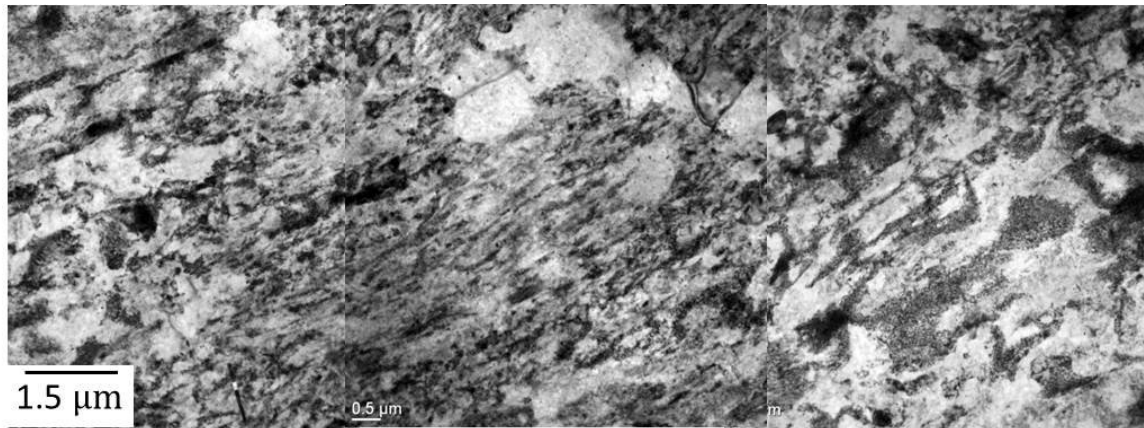


Figure 6-14 A montage of consecutive bright field TEM images from lesser strained regions at $T \approx 200\text{ }^{\circ}\text{C}$ and $\dot{\epsilon} \approx 5 \times 10^3/\text{s}$. Substantial dislocation activity is evident without any twinning.

EBSD is performed (same orientation with Figure 6.14) on one of the almost fully DRX microstructures (Figures 6.10, $\dot{\epsilon} = 1 \times 10^4/\text{s}$ and $T = 230\text{ }^{\circ}\text{C}$) representing the uniform flow behavior. Grains have high-angle grain boundaries and non-basal orientations where basal orientations (0002) are denoted by red color (Figure 6.15). This indicates that the initial ultrafine size DRX grains, which start out as low-angle and possess the shear type tilted deformation texture (non-basal texture) at the low deformation temperature, eventually grow and become high-angle as the temperature increased. Black regions in the EBSD picture, represents the low quality diffraction points. Reliable diffraction data was not possible in these regions due to high dislocation density and the residual strain. Therefore, black regions are a good indication of heavily dislocated microstructures that were observed in TEM.

Due to the limitations of the EBSD technique, the same analysis cannot be performed on the non-uniform microstructures where heavily dislocated regions occupy a large fraction of the structure. However, the same ultrafine grains in the non-uniform microstructures are believed to consume the dislocated regions at higher deformation temperatures and grow into the coarse, high-angle boundary grains shown in Fig. 6.15. In the meantime, dislocated regions in the lesser strained

areas also rearrange themselves into recrystallized grains as the dislocation cross-slip is promoted with increasing temperature. When the temperature is above $\sim 230^\circ\text{C}$ these two processes consume the entire non-uniform microstructure, transforming it into the fully DRX uniform one (Fig. 6.15). Therefore, the microstructure evolves continuously with the deformation temperature, where the high-angle DRX grains form continuously from the dislocated regions without the nucleation and growth of new strain-free grains (continuous recrystallization – CDRX).

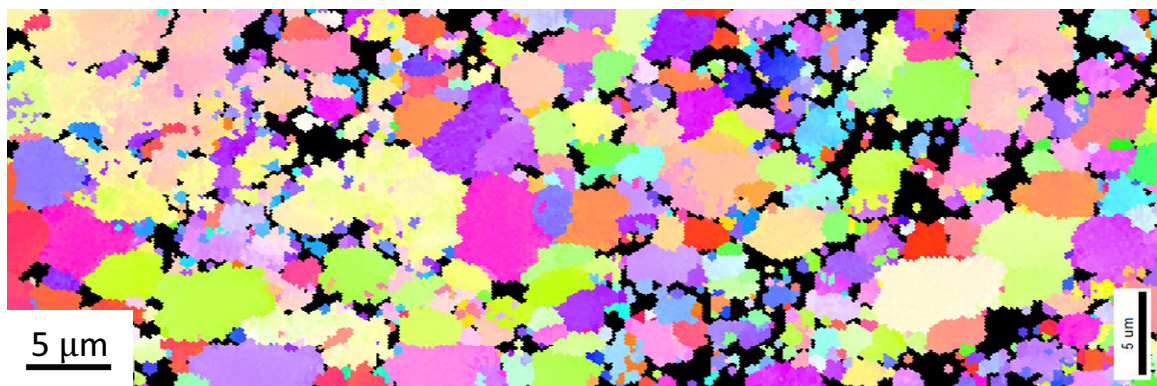


Figure 6-15 EBSD micrograph for the condition representing the uniform flow behavior, $T > 230^\circ\text{C}$. Grains have non-basal orientations (basal are denoted by red color).

Macro texture evolution with temperature presented in Figure 6.16 is in conjuncture with the continuous microstructural evolution. Unlike the typical basal texture observed in rolled sheets (same as the as-received material), in EM strips basal planes are tilted from the normal axis of the sheet and oriented approximately parallel to the shear plane. This non-basal texture is typical in shear-type deformation of magnesium [82, 93], and representative of the deformation texture. It is important to note that, the texture is essentially basal type but tilted along the shear direction. It is common to name this type of texture “non-basal” in the literature and “non-basal” nomenclature is also used in this thesis.

Non-basal texture is observed both at the low and high temperatures, again irrespective of the strain rate (Fig. 6.16). Therefore, the DRX grains at low temperatures have deformation textures and should keep this texture during their growth as they become high-angle type grains. If the microstructure evolution occurred by nucleation and growth of new strain-free grains, then the texture would have been basal type. In other words, the newly nucleated grains would have the same texture as their parent grains. However, basal texture component is absent in (0002) pole figures at all temperatures, suggesting that nearly all of the DRX grains carry deformation texture and formed by the dislocation and slip activity.

A spread in the non-basal texture with temperature can be observed from the (0002) pole figures. The same type of spread is also evident from the $(10\bar{1}0)$ pole figures, although the texture intensity is very minimal compared to the (0002) pole figures. There appears to be a rotation around c-axis at low temperatures based on the $(10\bar{1}0)$ pole figures, which may be related to the DRX process [82]. While it is hard to make conclusions on the type of the DRX process and the microstructure evolution from the $(10\bar{1}0)$ pole figures due to the weak intensity, (0002) pole figures clearly indicate that the tilt of the c-axis with respect to the chip normal direction remains unchanged with temperature. This still suggests that the deformation texture is persistent with the microstructural evolution which should be of continuous type.

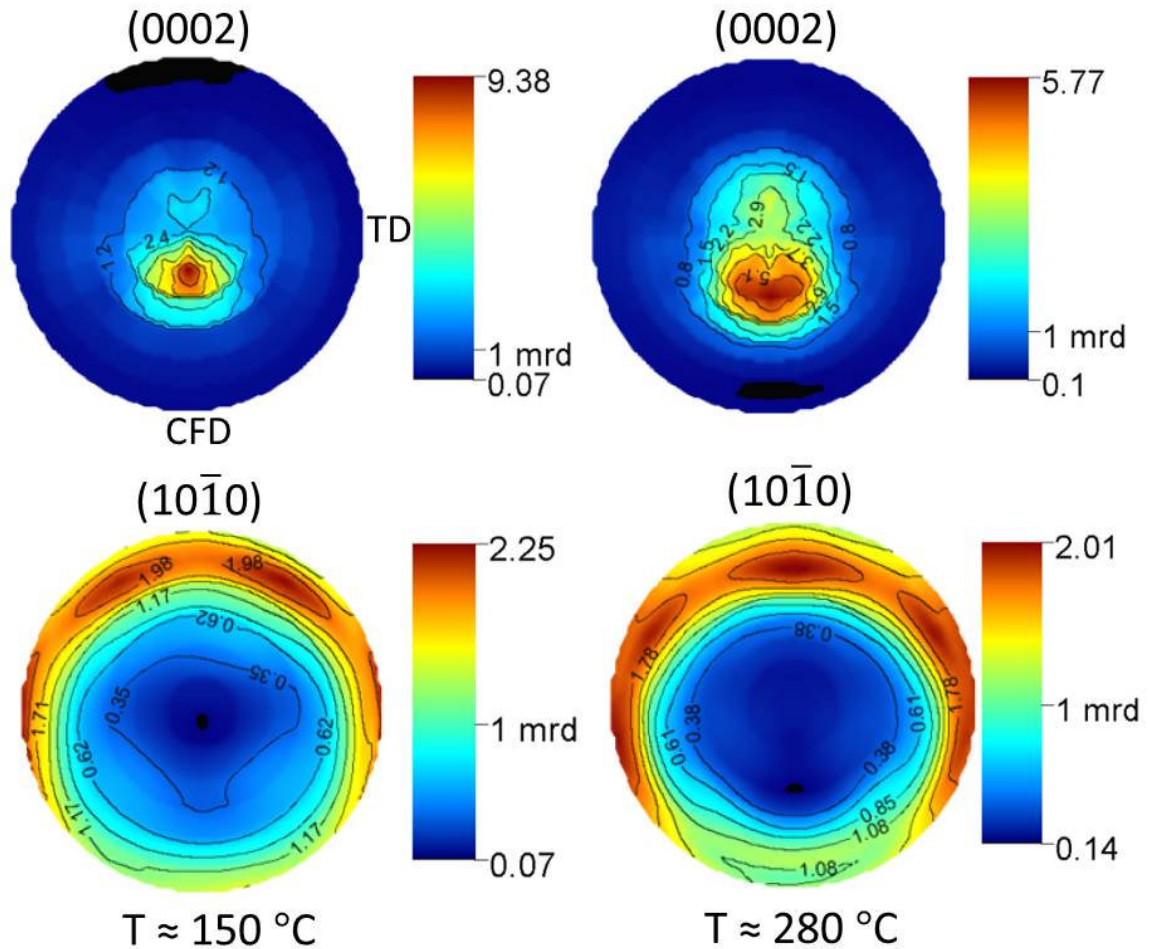


Figure 6-16 Macro texture evolution with the deformation temperature. (0002) and $(10\bar{1}0)$ pole figures are plotted for chip flow and transverse directions. The texture data is obtained from the center of the strips by mechanical polishing from surfaces.

The continuous type microstructure and texture evolution has been suggested before for magnesium (See Section 4.2). Continuous dynamic recrystallization (CDRX) is a type of DRX that can result in high-angle grain boundaries through dislocation accumulation, rearrangement and local lattice rotations given that the considerable cross-slip activity is allowed [78-81, 94-97]. Discontinuous dynamic recrystallization (DDRX), on the other hand, requires nucleation and growth of new grains, which requires dislocation climb activity [78, 80, 87] (Section 4.2).

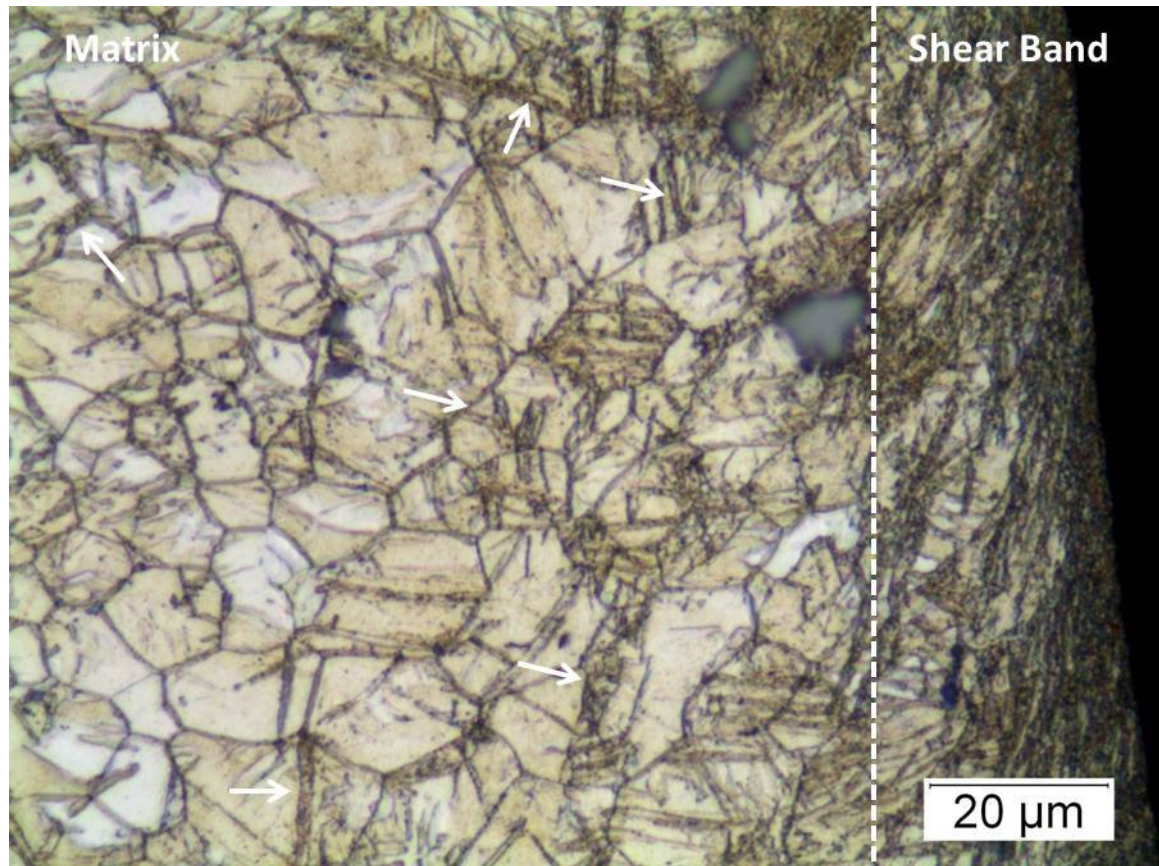
Dislocation climb is only possible at very low strain rates and high temperatures (Figure 4.3) and this is not the case for the EM experiments.

The cross-slip becomes favorable at high stacking fault energies. Although magnesium is known to have low stacking fault energy on basal planes ($< 50 \text{ mJ/m}^2$), the stacking fault energy increases considerably ($350\text{-}450 \text{ mJ/m}^2$) for partial dislocations on pyramidal and prismatic planes [98]. Therefore, pyramidal and prismatic slip should occur at high strain rates and temperatures with considerable cross-slip activity which then results in CDRX. Pyramidal and prismatic slip activity also known to increase with temperature (Figure 4.1). Although the temperature dependence of pyramidal and prismatic slip activity is reported at low strain rates (Figure 4.1), a similar temperature dependence is also expected at high strain rates due to the temperature rise effects. In other words, the non-basal slip activity is essentially a function of temperature (the deformation temperature at high strain rates), irrespective of the strain rate. Thus, the CDRX activity also is expected to increase with the deformation temperature. This supports our microstructural observations of increasing fraction of DRX area with the temperature. Then, the DRX regions in the microstructures are most likely the CDRX type.

Microstructure evolution with temperature in EM strips can be divided into three regimes:

- 1) At low temperatures, $80 < T < 150 \text{ }^\circ\text{C}$, twinning and basal slip occurs simultaneously. Twinning is mostly observed in lesser strained matrix. Some twins also recrystallize through the twinning-induced dynamic recrystallization (TDRX) mechanism albeit the fraction of TDRX regions being minimal (Figure 6.17). TDRX happens through the intersection of twins and can result in random (high-angle) boundaries. These boundaries can migrate easily and eliminate large number lattice dislocation which causes strain softening. TDRX regions can be a potential site for shear localization and initiation sites for adiabatic shear bands [78-81, 96]. Indeed,

TDRX is usually associated with flow softening observed in magnesium at low to warm temperatures and low strains [80].



T = 145 °C, LP EM

Figure 6-17 Twinning induced dynamic recrystallization (shown by white arrows) in LP EM strip processed at T = 145 °C.

Basal-slip dominates the shear banded regions where the strain is localized and high. This causes the highly non-uniform microstructures both on the local TEM scale (Figs. 6.13A&C) and the through thickness (optical microscopy) scale (Figure 6.10 – first microstructure from left and Figure 6.17). CDRX is possible even at these temperatures due to the high strain and hydrostatic pressures (Figure 6.13 B&D). Ultrafine (~ 150 nm) and equiaxed CDRX grains are observed. Rotational dynamic recrystallization (RDRX) can be suggested as a possible mechanism. RDRX is known

to occur with basal-slip near the original grain boundaries due to increased dislocation activity as a result of strain incompatibilities between adjacent grains [76, 77]. When the stress near grain boundaries exceed the CRSS required for non-basal slip, RDRX becomes possible. Thus, the initial grain boundaries will show RDRX leaving inside of the grains heavily dislocated that are mostly basal type [78, 81]. Moreover, activation of basal-slip is highly orientation dependent and only the initial grains that are oriented in favor of basal-slip will show considerable dislocation activity. Thus, fraction of CDRX regions is small at these temperatures due to the limited number of original grains showing dislocation activity (Figure 6.13 B&D).

2) At warm (intermediate) temperatures, $150 < T < 230$ °C, non-basal slip (prismatic $\langle 10\bar{1}0 \rangle$ and pyramidal $\langle c+a \rangle$) dominates the flow. Non-basal slip is reported to occur in magnesium when $T > 250$ °C [78, 80, 99]. High strains and hydrostatic pressures are shown to decrease this transition temperature [40, 96]. The fraction of CDRX regions increase in this regime with enhanced cross-slip, since dislocation accumulation and rearmament is possible with cross-slip (Figure 6.14 and Figure 6.13E). Moreover, 3D recrystallization requires $\langle c+a \rangle$ dislocations. Cross-slip becomes easier when non-basal systems are active due to the high surface energy of these systems.

Microstructures on the thickness scale (optical microscopy) also becomes more uniform with the extent of CDRX increasing, characterized by weak shear localization and vague shear bands (Figure 6.9 – second and third pictures from left). While CRSS of basal slip does not depend on temperature (Figure 4.1), CRSS of non-basal slip is highly dependent and decreases sharply with temperature (Figure 4.1). The sharp decrease occurs near 200 °C. Therefore, CDRX is expected to occur widely in the microstructure, replacing the basal slip. Initially, low-angle grain boundaries form (dislocation structures). Then, trapping of dislocations by low-angle boundaries increase their misorientation. These boundaries also rotate to

accommodate more strain and they will eventually evolve into high-angle boundaries. This process usually results in the steady-state flow stress [78].

3) At high temperatures, $230\text{ }^{\circ}\text{C} < T < 300\text{ }^{\circ}\text{C}$, CDRX happens almost everywhere in the microstructure (Figure 6.15 and Figure 6.13 G&H). DDRX event is not possible due to high strain rates and diffusion type processes are limited [78, 95, 97]. Non-basal slip becomes even more easier with further increase in the temperature. It even happens in the initial grains which were not oriented favorably for the slip. Non-uniformity in the microstructure is reduced and CDRX grains get bigger in size ($0.5 - 1\text{ }\mu\text{m}$) and relieve more strain. The strips also become completely uniform on the thickness scale devoid of shear bands or any form of localization (Figure 6.9 – fourth picture from left and Figure 6.10). Grains have high misorientation angle and non-basal deformation textures at this stage (Figure 6.15).

The temperature regimes defined here are applicable to all kinds of shear deformation processes (such as ECAP, DSR, UCAP and Conform-ECAP [93, 100]) as long as the effective strain and hydrostatic pressures are similar. The temperatures may shift to higher values ($+ 30 - 50\text{ }^{\circ}\text{C}$) as the effective strain and/or hydrostatic pressure decreases. This is based on our observation on LP EM conditions. The LP EM strips exhibited similar mechanisms compared to HP EM with a critical transformation temperature of $\sim 260\text{ }^{\circ}\text{C}$ for uniform flow with complete CDRX (Figures 6.5 and 6.11).

6.2.5 Effects of Recrystallization Mechanisms on Shear Banding

Recently, Rittel and coworkers [84] suggested that DRX event as the main reason for the shear banding. They identified the dynamic stored energy of cold work as the main driving force for shear band formation and developed an alternative microstructural based model to the classical mechanistic models for shear band prediction [85]. According to their findings, stored energy of cold work is the fraction of the mechanical energy that remains in the microstructure through dislocation rearrangements. The shear banding and instability becomes less likely,

as the deformation energy gets stored in the microstructure. Strain-free and ultrafine DRXed grains introduce perturbations in the microstructure as they become “soft” inclusions around their “hard” heavily dislocated surroundings and thereby reduces the stored energy in the microstructure [84].

They observed very similar TEM microstructures to the ones presented in Figures 6.13 and 14. In this work, non-uniform microstructures that are mixtures of ultrafine strain-free grains and heavily dislocated regions are also associated with shear banding. However, the tendency for shear banding decreases as the extent of the CDRX event increases. Uniform flow is possible when the microstructure is fully DRX (Fig. 6.13F). The reason of the non-uniformity in the microstructure is believed to be the limited cross-slip activity. Unlike the DRX regions, heavily dislocated regions may lack enhanced slip activity due to their non-favored orientations. As temperature increases, cross-slip activity is increased and the CDRX event becomes possible throughout the microstructure. When the critical temperature is realized either by increasing the strain rate or the pre-heating temperature, shear banding is not possible, even if the microstructure is still a mixture of CDRX and heavily dislocated regions (Fig. 6.13G). Although the fraction of the latter becomes minimal when the critical transition temperature is realized.

Fraction of the CDRX regions increases with non-basal slip activity. When non-basal slip (and cross-slip) becomes the dominant deformation mechanism, strain-rate sensitivity increases. Steady-state flow behavior observed with strain (i.e. no flow softening), indicating effective accumulation and rearrangement of dislocations through cross-slip. As strain-rate sensitivity becomes high, any localization to CDRX regions will not be possible even if they act as soft spots in the microstructure. Then, heavily dislocation regions also recrystallize and result in uniform microstructures.

However in low temperature deformation, strain rates sensitivity will be small due to the dominant deformation mechanism of basal slip and twinning. Initiation of the

low temperature forms of CDRX, TDRX/RDRX, at these temperatures relieves strain and nucleates soft spots in the microstructure. Shear localization to the TDRX/RDRX regions is now possible due to the low strain-rate sensitivity. Strain and strain rate will increase in the TDRX/RDRX regions due to the localization, resulting in less strained matrix and DRXed shear bands. Even inside the shear bands, the microstructures is non-uniform on the local level (Fig 6.13A&C).

Indeed, CDRX and DDRX are shown the increase the strain-rate sensitivity in magnesium alloys [86]. Prasad and co-workers have developed processing maps to study the workability and deformation behavior of metals at hot working ($0.5 T_m$) temperatures. They established a similar continuum microstructural model that is similar to Rittel's, which is based on the power dissipation through microstructural mechanisms. A non-dimensional efficiency index given as:

$$\eta = \frac{2m}{m + 1} \quad (54)$$

is used to measure the power dissipation through microstructural processes. The relation is theoretic, where m is the strain-rate sensitivity. Workability increases as efficiency index increases since more power is dissipated [69, 101]. Processing maps are also developed for both cast and wrought forms of magnesium AZ31 [86, 102]. Although the range of temperature and strain rate in the processing maps is not relevant to the thermo-mechanical conditions in EM, the CDRX and DDRX mechanisms are shown to have high ($\sim 40\%$) efficiency index due to the increased m value. Therefore, hot working without any instability was possible with these mechanisms.

6.2.6 Modeling of the Shear Banding Behavior

Shear localization models derived in sections 3.1 and 3.2 will be used in this section to model the adiabatic shear banding behavior in magnesium. It was shown that susceptibility to shear banding is calculated for a shear-type deformation through the following equation (Eqs. 38 and 44 in Chapter 3):

$$\alpha_m = -\frac{\sqrt{3}}{m} \left\{ \frac{1}{\tau} \left(\frac{\partial \tau}{\partial \dot{\gamma}} \right)_{\dot{\gamma}, T} + \frac{\beta_1}{\rho C_p} \left(\frac{\partial \tau}{\partial T} \right)_{\dot{\gamma}, \dot{\gamma}} \right\} \quad (55)$$

which was same for both load instability (Eq. 38, α_f) and perturbation analyses (Eq. 44, α_m) models. A similar equation for susceptibility can also be derived for effective quantities

$$\alpha_m = -\frac{1}{m} \left\{ \frac{1}{\sigma} \left(\frac{\partial \sigma}{\partial \dot{\epsilon}} \right)_{\dot{\epsilon}, T} + \frac{\beta_1}{\rho C_p} \left(\frac{\partial \sigma}{\partial T} \right)_{\dot{\epsilon}, \dot{\epsilon}} \right\} \quad (56)$$

where m is the strain-rate sensitivity (same for both shear or effective quantities – Eq. 37). Using the power law relationship for flow stress;

$$\sigma = \sigma_0 \dot{\epsilon}^n \quad (57)$$

the first term in Eq. 56 can be simplified to the strain-hardening rate:

$$\frac{1}{\sigma} \left(\frac{\partial \sigma}{\partial \dot{\epsilon}} \right)_{\dot{\epsilon}, T} = \frac{n}{\dot{\epsilon}}; \quad (58)$$

and the second term in Eq. 56 ($\partial \sigma / \partial T$) is the flow softening rate.

Assuming constant material properties over a range of thermo-mechanical conditions, which was shown to be erroneous in magnesium (Section 4.2), shear banding models suggest that the more heat generated from deformation and more of that heat transferred to chip, thus increasing adiabaticity, the more will be the susceptibility to flow softening and localization. In other words, the shear banding theory suggests that temperature rise in the deformation zone causes adiabatic shear banding.

The flow softening term in Equations 55 and 56 is multiplied by both the fraction of the deformation energy transferred to heat and the fraction of this heat staying in the deformation zone. Shear banding becomes possible when the conditions become more adiabatic. Then the heat partition coefficient, β_1 , is an indication of adiabaticity. Heat partition coefficient calculated by Eqs. 19 and 20 is plotted in Figure 6.18 as a function of deformation rate and initial chip thickness for magnesium AZ31. The

coefficient saturates and approaches to 1 as deformation rate gets closer to 4 m/s. An increase in the initial chip thickness will also result in more adiabatic conditions.

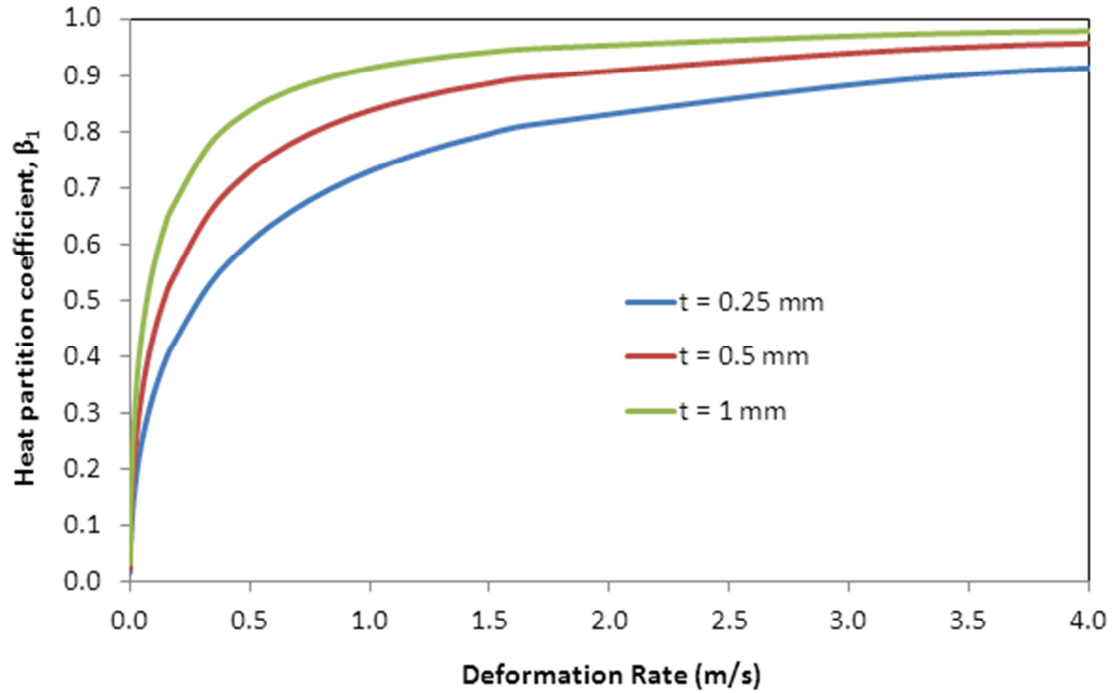


Figure 6-18 Heat partition coefficient vs. deformation rate at various initial chip thicknesses (feed). Calculations are for magnesium (Equations. 19 and 20).

Eq. 56 is plotted at an effective strain of 1.15 with deformation rate assuming constant material properties in Figure 6.19 – blue curve. Strain-rate sensitivity is taken as 0.02 (low temperature deformation of magnesium) [71], strain-hardening rate is 0.15 [70] at 25 °C. The flow softening rate is -0.75 MPa/K [67]. Of course, all the material properties quoted above are obtained at low effective strains (~ 0.1) and low strain rates ($< 10^{-1}/s$).

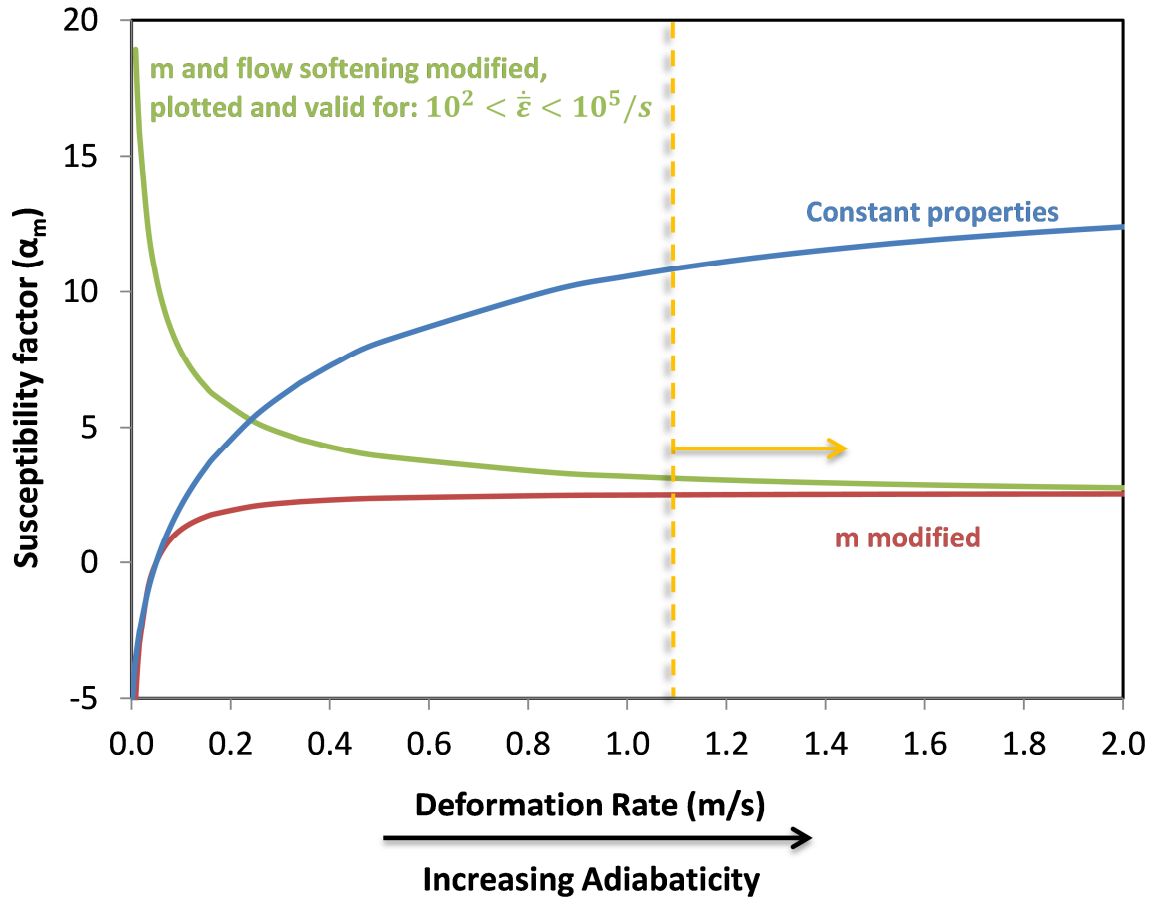


Figure 6-19 Susceptibility factor vs. deformation rate. Classical shear banding behavior with constant material properties (blue – Eq 56), stabilization effects of temperature-dependent m on susceptibility (red – Eq 56 with temperature dependent m) and m and flow softening modified (green - Eq 59 with temperature dependent m) curves are given. Flow softening and m -modified susceptibility curve (green) levels-off around 1.1 m/s.

Susceptibility to shear banding increases and becomes greater than 0 (a necessary condition for shear localization) with the deformation rate (Fig. 6.19 – blue curve). Therefore, at a constant initial chip thickness, an increase in the deformation rate (or adiabaticity) should result in transformation from continuous strips to the shear banded and segmented strips. This might have been true for other materials such as titanium and stainless steel [4, 12, 13]; but it is certainly not applicable to magnesium, which shows a marked change in flow properties with temperature. The exact opposite behavior is observed in Mg AZ31. Materials like titanium and

stainless or alloy steels may have relatively constant flow properties and may have tendency to segment as the deformation rate is increased. Magnesium, on the other hand, shows a considerable increase in plasticity through the switch in the deformation mechanisms, which then changes the flow properties.

Indeed magnesium AZ31 exhibits a sharp increase in strain-rate sensitivity with temperature (Figure 4.5). Therefore, the stabilizing effects of strain-rate sensitivity can be realized (Equations. 55 and 56) as temperature is increased with deformation rate. In order to demonstrate the stabilizing effects of strain-rate sensitivity, the susceptibility to shear banding is plotted (Eq. 56) by making strain-rate sensitivity dependent on temperature through the temperature rise created by increasing deformation rate (Fig. 6.19 – red curve). Figure 4.5 is used to estimate the temperature dependency of m . Again, m is assumed to be independent of the strain rate, thus deformation rate itself has no direct effects. Susceptibility plateaus around 125 °C (0.1 m/s), due to the marked increase in m (Fig. 6.19 – red curve).

While the m -modified-susceptibility curve (red) predicts the lack of shear banding at high deformation rates and temperatures, low temperature shear banding behavior still needs to be modeled. Thus, the strain-hardening rate and flow softening rate data should be obtained. The following equation can be used to simplify these two effects by using Equations 34-37:

$$\alpha_m = -\frac{\sqrt{3}n'_\tau}{m} = -\frac{\sqrt{3}\left(\frac{\partial \ln \tau}{\partial \dot{\gamma}}\right)_{\dot{\gamma}}}{m} \quad \text{or} \quad \alpha_m = -\frac{n'}{m} = -\frac{\left(\frac{\partial \ln \sigma}{\partial \dot{\epsilon}}\right)_{\dot{\epsilon}}}{m} \quad (59)$$

where n' and n'_τ are flow hardening/flow softening rates [15, 63]. The flow hardening/flow softening rate can be obtained from stress-strain test data that is not corrected for deformation heating at constant strain rate. In other words, test data that intrinsically contain the effect of deformation heating on flow hardening/flow softening rate [15].

For example, hot compression test data having a similar level of adiabaticity to ECAP can be used to determine n' when modeling the shear banding behavior [15, 63, 103]. In order to successfully model the deformation process, it is important that the mechanical test has a similar level of adiabaticity -“heat partition coefficient”- to the deformation process at a given deformation rate.

For the case of EM, quasi-static and dynamic shear test data obtained from hat-shaped specimens can be used [104]. El-Magd and Abouridouane showed that both quasi-static ($10^{-2} - 10/s$) and dynamic ($10^2 - 10^5/s$) shear flow curves exhibit the same n'_τ at a shear strain of 2 ($\epsilon = 1.15$). Therefore, it is fair to assume a constant flow hardening/flow softening rate over a wide ranges of deformation rates at a fixed shear strain of 2. The m-modified-susceptibility curve is re-plotted for a constant n' of -0.27 (Figure 6.19 – green line). This value is obtained with the previous material property data (i.e. $n = 0.15$ and $(\partial\sigma/\partial T) = - 0.75$ MPa/K) at the maximum strain rate, where adiabaticity was maximum. Shear banding now becomes possible at low strain rates, $\sim 10^2/s$, and temperatures. The susceptibility factor decreases and levels-off around 1.1 m/s due to the stabilizing effects of strain-rate sensitivity with temperature. The m-modified (red line) and m-flow-softening-modified (green line) curves merge at high strain rates since the adiabaticity is maximum and they both have the same n' of -0.27.

The m-flow-softening-modified susceptibility curves are plotted for a wide range of flow hardening/flow softening rates including the data obtained from El-Magd and Abouridouane (Figure 6.20). They predicted a rate of -0.6 at $\gamma = 2$ and $T_0 = 25$ °C for $10^2 < \dot{\gamma} < 10^5/s$. Assuming their test and EM have the same degree of adiabaticity, this value can be directly used for modeling the shear banding behavior since the deformation conditions in EM fits in the thermo-mechanical conditions tested.

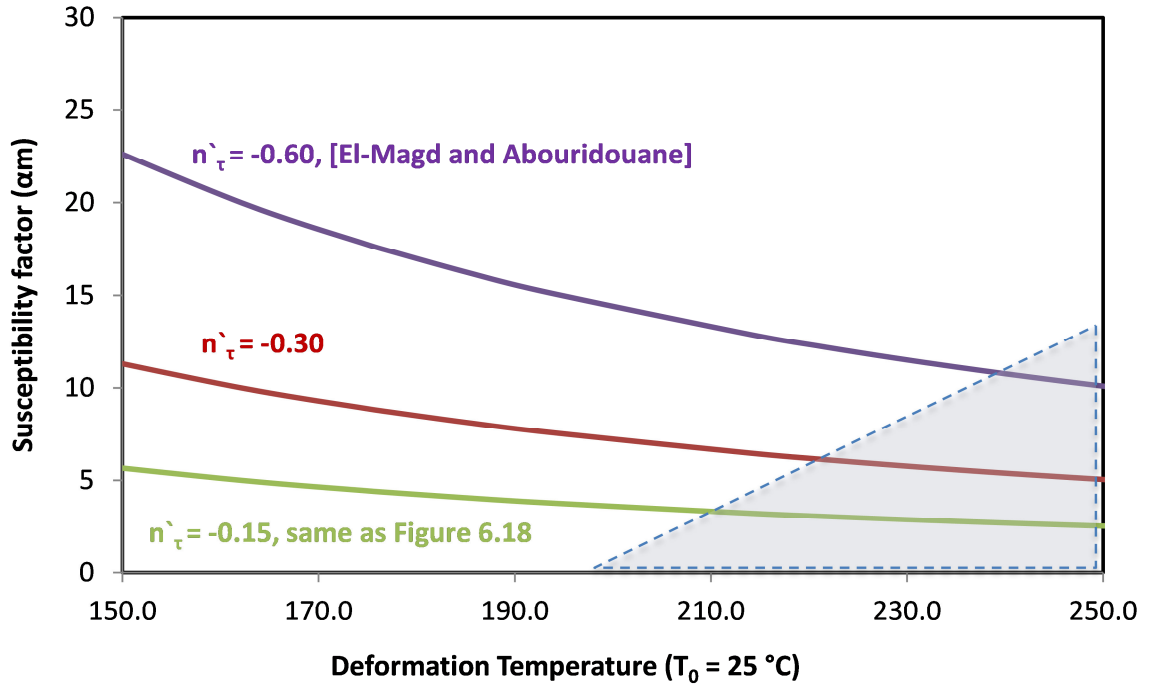


Figure 6-20 Susceptibility factor vs. deformation temperature. Various flow hardening/flow softening rates are plotted. Dotted triangle covers the temperature ranges where susceptibility curves level-off relatively.

However, based on the observations from EM experiments, $n'_\tau = -0.30$ ($n' = -0.52$) seems a better fit for modeling. Susceptibility to shear banding at $n'_\tau = -0.30$, levels-off around 220 °C and decreases below the value of 5. This value is quoted as a threshold number above which catastrophic shear banding is observed in various metals [4]. An increase in flow hardening/flow softening rate, increases the critical transition temperature above which shear banding will not be possible. The dotted triangle in Figure 6.20 covers the temperature range for uniform deformation for various flow hardening/flow softening rates.

The susceptibility factor is very sensitive to the deformation rate for low strain-rate sensitivity. As the sensitivity increases, flow localization is stabilized and does not depend on the adiabatic temperature rise. As discussed before, α_m should be more than 3-5 for catastrophic shear banding. Therefore, there is a critical temperature

where a switch in deformation behavior occurs. Beyond this critical temperature, m is so high that α_m levels-off.

Therefore, the switch between segmented chips to continuous chips can be explained by the theory provided that the material properties are highly dependent on the thermo-mechanical conditions. It is now evident that magnesium AZ31 changes its deformation behavior with the application of sufficient temperature. This change in the deformation mechanism suppresses shear localization.

The flow hardening/flow softening rate depends on the hydrostatic pressure and the strain. It will increase with decreasing hydrostatic pressure which then increases the critical transition temperature for uniform flow.

6.3 Summary for Mg AZ31 Study and Processing Map

Continuous strips of Mg AZ31B have been produced directly from the material in bulk form, in a single step extrusion-machining operation with minimal or no pre-heating of the workpiece. The alloy, which is inherently susceptible to shear banding, becomes workable by introduction of high hydrostatic pressures, $p/2k = 1.7$ and high temperature (~60% of melting temperature) arising from local deformation heating. The hydrostatic pressure and temperature in the deformation zone are modeled; and the models are used to perform a study of the effects of these parameters on the chip formation process, by varying one independently from the other. This study shows in addition to elevated deformation zone temperatures, the alloy is most workable when the applied hydrostatic pressure is high. Increasing the feed rate complicates workability. Larger thickness strips can be achieved, however, by increasing the hydrostatic pressure or the deformation zone temperature by initial pre-heating. Reducing the chip compression ratio promotes both higher pressure and temperature, thus enabling production of larger section sheets with minimal pre-heating.

The process is capable of producing homogeneous microstructures, showing considerable grain refinement from ultrafine (~ 150 nm) to very fine grain sizes ($0.5 - 3 \mu\text{m}$) together with non-basal textures which are both essential in subsequent forming operations. Increased grain refinement (and strength) is achieved by reducing the temperature in the deformation zone necessary to produce continuous chip formation (i.e., applying larger hydrostatic pressure). A uniform microstructure is achieved by increasing this pressure, as it results in elimination of shear localization.

The shear banding models are revised and the shear banding is concluded to be primarily a function of the deformation temperature. A critical deformation temperature is established above which shear banding becomes not possible. This temperature depends on the effective strain and hydrostatic pressure. An increase in pressure and/or strain decreases the critical transition temperature.

A direct relationship between fraction of DRX and susceptibility to shear banding is found due to the temperature dependency of the deformation mechanisms. The extent of DRX increase with the cross-slip activity, which was shown to be a function of temperature. When DRX (especially CDRX) happens uniformly throughout the microstructure with the aid of cross-slip, shear does not localize to the DRX regions.

The shear banding behavior together with deformation and DRX mechanisms is summarized in the following processing map in three temperature regimes (Figure 6.21). The deformation temperature can be realized either by increasing the strain rate or the initial temperature. The increasing strain rate has minimal effects on the deformation behavior at a constant deformation temperature, thus the map below is valid for all the strain rates used in this study.

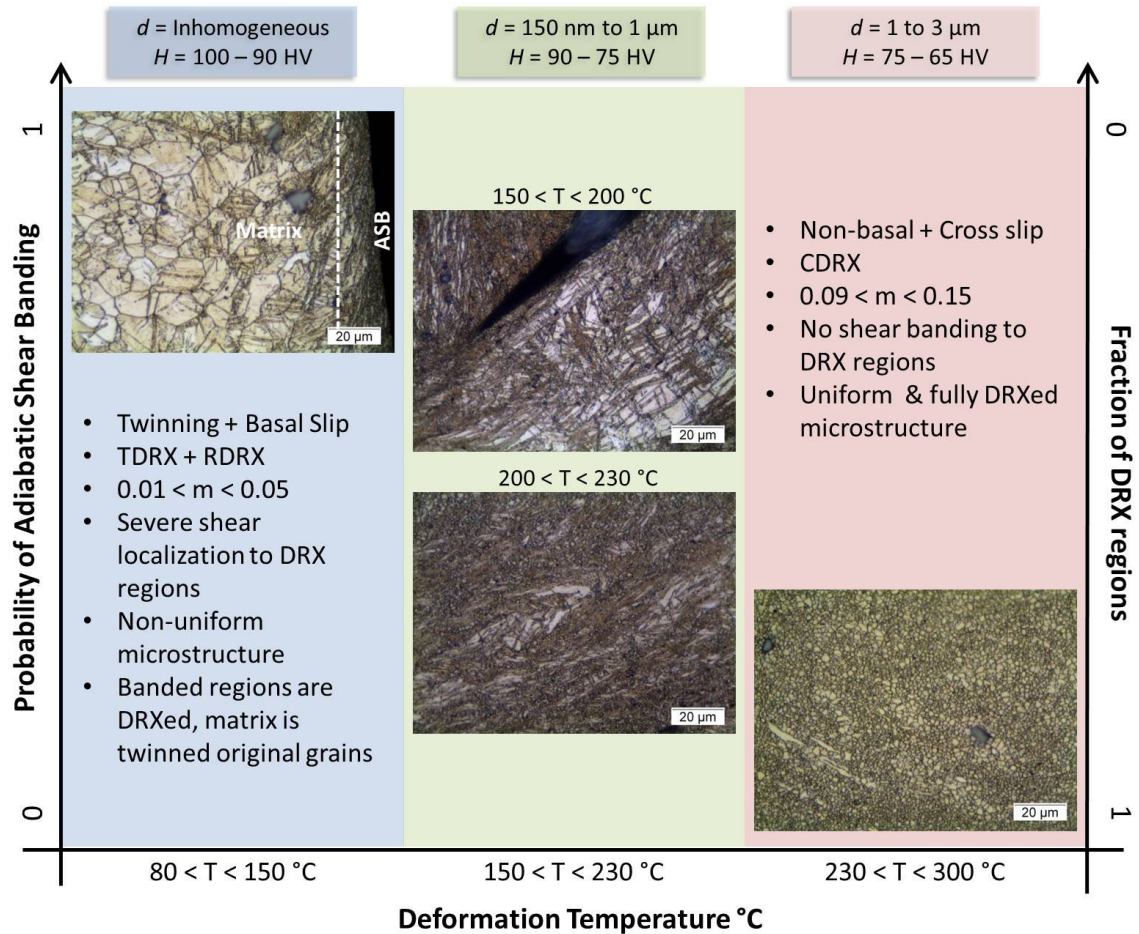


Figure 6-21 Processing and deformation mechanism map of magnesium AZ31 for $10^3 < \dot{\epsilon} < 10^5/\text{s}$ and $1.08 < \epsilon < 1.22$. Blue and red areas indicate low and high temperature deformation respectively. Deformation mechanisms and the shear banding behavior are a mixture of two at intermediate (warm) temperatures (green).

The absolute values of the temperatures presented in Figure 6.21 are of course valid for the thermo-mechanical conditions tested in this study. While the absolute values of the temperature regimes will be different for various processes, one can still expect the same trend in the shear banding behavior and microstructure evolution (i.e. fraction of DRX) with the temperature. The three temperature regimes will still be observed with temperature values being specific to the selected processing conditions. However, it is possible to obtain the temperature values from the modified shear banding model (Figure 6.20). Once the flow hardening/flow

softening rate is measured through simple mechanical tests, the critical transition temperature can be established. Then, the other temperature values can be updated based on the transition temperature.

The microstructure and shear banding evolution in a given temperature regime can be divided into the following steps:

- 1) The deformation temperature determines the operating deformation mechanism.
- 2) The dominant deformation mechanism determines the type of DRX.
- 3) Fraction of the DRX regions is determined by the temperature. When non-basal and cross-slip activity increases with temperature, DRX starts to occur uniformly throughout the microstructure. This also increases the strain-rate sensitivity, m , which depends on the deformation mechanism, thus the deformation temperature. The fraction of DRXed regions also increases and recrystallized grain size decreases with increasing hydrostatic pressure (and strain) due to enhanced cross-slip activity.
- 4) If the fraction of the operating DRX is below a critical value, strain localizes in these “soft spots” (TDRX/RDRX) regions in the microstructure. If the DRX mechanism operates everywhere in the microstructure due to the increased cross-slip activity, such localization is not possible to the DRX (CDRX in this case) regions.
- 5) At low temperatures, strain further localizes to TDRX/RDRX regions if m is smaller than 0.05, as it cannot stabilize the localization. Catastrophic shear banding occurs with cracking and segmentation along the shear bands. At high temperatures, m stabilizes the localization if such an initiation of shear banding is possible. In other words, even if shear starts to localize to the CDRX regions, m prevents further localization and shear banding and DRX will uniformly continue throughout the microstructure.
- 6) Uniformity of the macrostructure (microstructure on the thickness scale) is determined by the shear banding activity. Shear bands result in the less strained matrix, which may consist of the original grain structure with little deformation.

CHAPTER 7. APPLICATION OF EM TO ALLOYS WITH LIMITED WORKABILITY – CAST AND SINTERED MATERIALS

7.1 Sintered Copper

Sintered pure copper is an ideal material to study the effects of porosity on the segmentation behavior. It has ~ 5% porosity that is homogenously distributed. The pores are very fine and closed, i.e. not interconnected. The matrix does not contain any other imperfections or impurities. The deformation mechanisms of pure copper without the porosity is well understood and copper has excellent workability with high strain-hardening rate and without any adiabatic shear band formation due to its high thermal diffusivity – $1.18 \times 10^{-4} \text{ m}^2/\text{s}$. Thus, it is not susceptible to shear banding as adiabatic deformation cannot be realized. Therefore, sintered copper can be used to model cast materials that are not susceptible to ASB.

Figure 7.1 shows the microstructure of sintered copper. Pores are homogeneously distributed throughout the microstructure and average size is smaller than $5 \text{ }\mu\text{m}$. Grain size ranges between $10 - 20 \text{ }\mu\text{m}$.

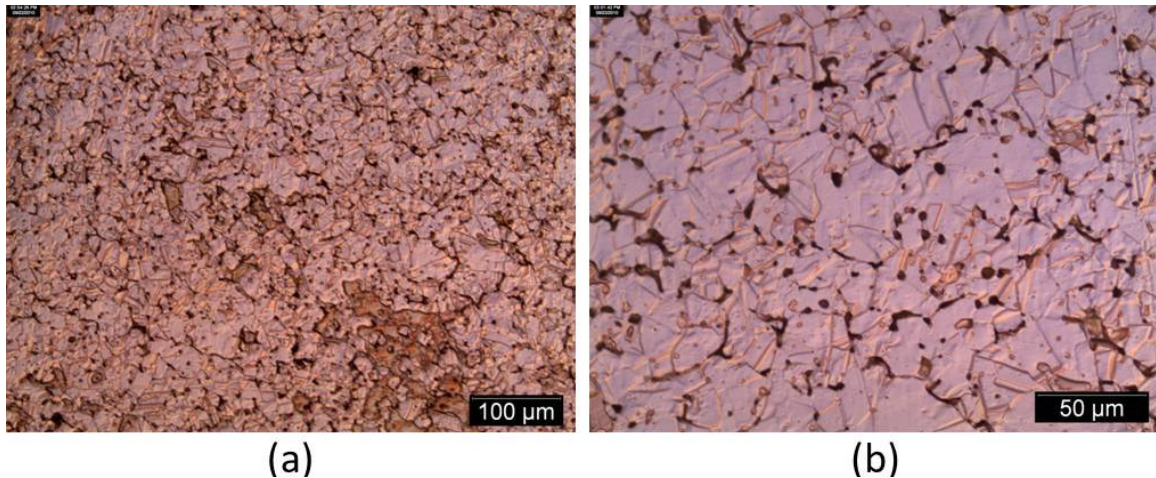


Figure 7-1 Optical microstructures of pure sintered copper. (a) Low magnification images shows the distribution of porosity. (b) High magnification shows the size of pores with grain structure.

Extrusion machining is applied to the sintered copper with 0.7, 1.6 and 1.8 chip thickness ratio. The initial chip thickness and deformation rate was kept constant at: 0.25 mm and 1 m/s, respectively. One of the important problems in EM of sintered copper was side cracking, especially for lower λ s. Although the extent of cracking was minimal - about 10% width -, this problem is solved by the application of side constraints on the EM tool. This is achieved by designing and manufacturing an EM tool with side walls. The effects of hydrostatic pressure and strain on the strip formation and pore closure behavior is investigated by changing the chip thickness ratio, λ and rake angle, α , simultaneously. Application of hydrostatic pressure by various chip thickness ratios was hypothesized to result in pore closure and relative density increase. Figure 7.2 summarizes the strip formation and pore closure behavior for a range of hydrostatic pressures and strain. Continuous strips are obtained for the processing conditions listed in Figure 7.2, with the exception of minor serration for $\lambda = 1.8$.

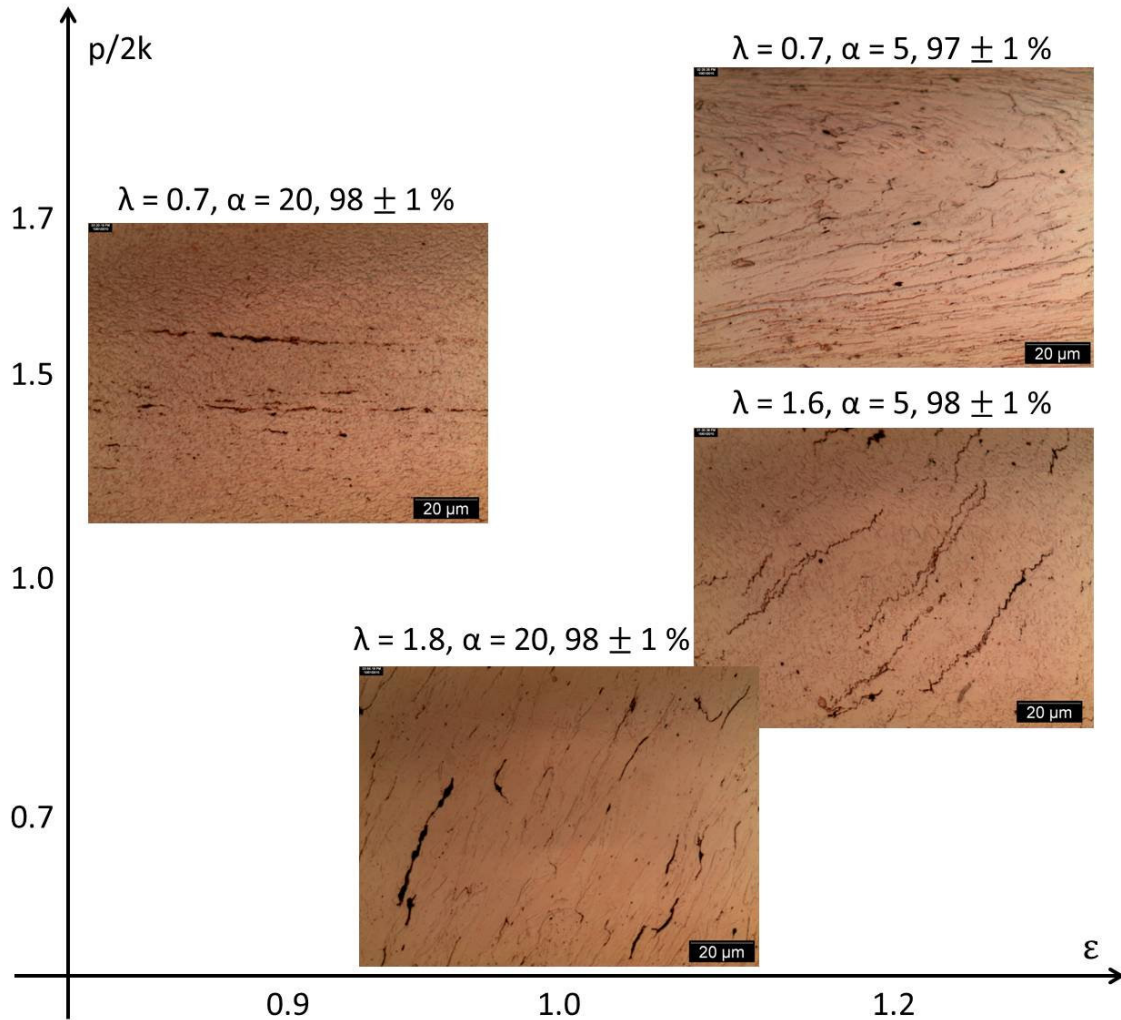


Figure 7-2 Optical microstructures of EM sintered copper strip cross-sections showing pore closure behavior at different levels of hydrostatic pressures and strains. The relative densities are indicated on top of each condition together with the processing conditions. $T_0 = 25\text{ }^{\circ}\text{C}$ and $V = 1\text{ m/s}$ for all the strips.

All of the EM tests were done at room temperature. Pre-heating made copper extremely malleable and the material did not go through the extrusion machining tool opening. Instead, it accumulated in front of the opening resulting in side flow. Room temperature EM resulted in good quality strips without any segmentation. EM strips densify slightly to $98 \pm 1\%$ relative density. The bulk was measured to be $96 \pm 1\%$ (all density measurements by Archimedes). Pore closure is observed at high hydrostatic pressures together with some remnant elongated and smeared pores.

The direction of elongation is almost parallel to chip flow direction (CFD). EM strips at low pressures also show pore closure, but the fraction of the remnant elongated and smeared pores increase slightly with lower pressure. Strain (in the tested range) does not seem to change the densification behavior. Elongated pores are aligned parallel to the shear plane and they are tilted from the chip flow direction. All microstructures are severely deformed and grain size is not resolvable under optical microscopy due to ultrafine grain formation. Segmentation and cracking are not observed at any of the processing conditions and pores do not interact with each other and does not become pathways for easy crack initiation and growth. Low magnification images in Figure 7.3 show the uniform deformation behavior with pore distribution and densification. Densification behavior becomes evident and clear when high magnification images in Figure 7.3 is compared with the bulk in Figure 7.1b.

An ideal way of assessing the pore closure behavior would be the tensile test of the strips along the chip flow direction. The real difference between the pore closure and elongation behavior for higher and lower λ values can be observed after the tensile deformation. Pores will open after the tensile deformation if the metallurgical bonding between the pores surfaces is weak, as expected for the case of higher λ values. At higher hydrostatic pressures and strains, metallurgical bond is expected to be stronger. Pore surfaces are expected to weld and should not open after the tensile test. Therefore, tensile testing of the strips is strongly recommended as a future work.

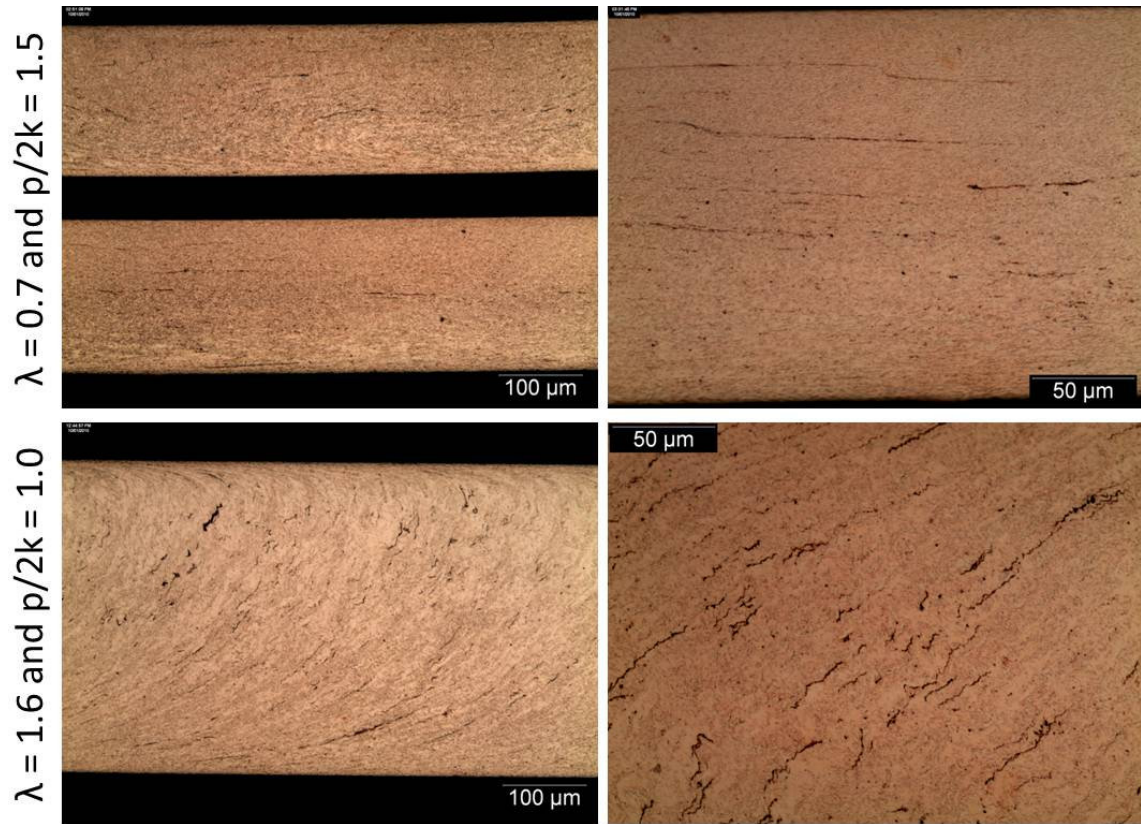


Figure 7-3 Representative optical images showing the uniform deformation (i.e. no segmentation and cracking) and densification behavior at (top) $\lambda = 0.7$ - higher pressure and lower strain, $\varepsilon = 0.9$, and (bottom) $\lambda = 1.6$ - lower pressure and higher strain, $\varepsilon = 1.2$. $T_0 = 25^\circ\text{C}$ and $V = 1\text{m/s}$ for both conditions.

7.2 Cast Brass 360

Brass 360 alloy, a.k.a. free machining brass, is specifically designed for easy machinability, meaning that discontinuous chips are easily obtained for best surface finish. This alloy is used for parts that require significant machining after the casting. Powder-like chips are readily formed during the machining of this alloy, resulting also in excellent machinability and increased productivity.

These brasses are intentionally alloyed with lead to create weak lead pockets in the microstructure. Lead has almost no solubility in brass and it precipitates as lead

pockets during the solidification. Cracks can easily initiate and grow from these pockets, resulting in powder-like chip formation during machining. This increases productivity in machining via faster machining speeds, longer tool life and excellent surface finish. Therefore, it is extremely difficult to get continuous chips in this alloy. Even though the continuous forms of this alloy will have no commercial application, application of EM to this alloy is ideal for studying the effects of weak second phases on the segmentation behavior. Furthermore, this brass is a good model material to test the capabilities of EM with alloys having limited workability, as it bears the greatest challenge among all.

Table 7.1 shows the chemical composition of the cast alloy provided by the manufacturer. Figure 7.4 is the as-cast microstructure consisting of dendrites, porosity and lead pockets. Porosity seems to be finely distributed through the microstructure. Hardness is measured to be 82 ± 3 HV.

Table 7-1 Chemical composition (wt. %) of copper alloy C-36000 (ASTMB-16).

Cu	Pb	Zn	Sn	Fe
61.45	3.32	34.75	0.16	0.25

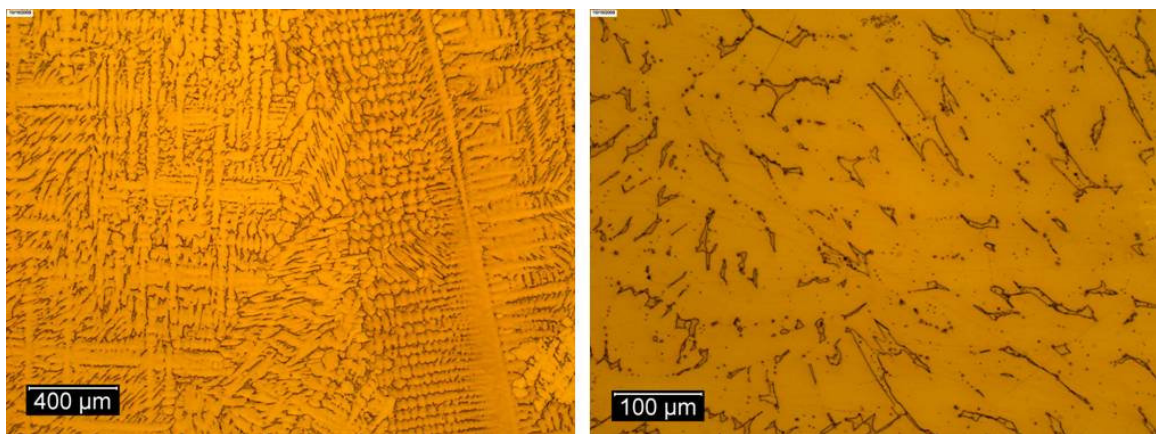


Figure 7-4 Optical micrographs of the as-cast brass 360 alloys. Dendritic solidification is observed with porosity and fine lead pockets (black dots in high magnification image).

Table 7.2 lists the multiple trials in order to get the continuous strips. LP EM (peeling with constraining tool barely touching the top surface) did not result in any success. Then, HP EM was tried at three different chip thickness ratios: 0.3, 0.5 and 1. Pre-heating is almost always necessary for the continuous strips with minimal cracking, although decent quality continuous strips can be obtained without pre-heating. Hydrostatic pressure is again essential in suppressing segmentation and cracking. An ideal chip compression ratio of 0.5 is obtained for the best quality strips resulting in minimum side and internal cracking. Side cracking in $\lambda = 0.3$ strips can be overcome by addition of side walls to the EM tool. Deformation rate has no significant effect on segmentation, once the critical rate is passed and adiabatic conditions are generated. It is possible to decrease the pre-heating temperature by decreasing the strain through an increase in rake angle at constant chip thickness ratio.

Table 7-2 EM tests on 360 brass and processing conditions. HP EM conditions and respective quality of the strips are color coded. Higher λ is marked with darker pink color and higher quality is marked by darker blue color. $t_0 = 0.3$ mm for all the experiments.

Rake angle/chip compression ratio	Deformation Rate (m/s)	T_0 (°C)	Appearance
+5° (peeling)	1	ambient	small pieces
+5° (peeling)	2	ambient	small pieces
+20° (peeling)	1	ambient	small pieces (longer than with 5°)
+20° (peeling)	2	ambient	small pieces (longer than with 5°)
+5°/0.3 (EM)	1	ambient	longer pieces with many side cracks
+5°/0.3 (EM)	2	ambient	longer pieces with many side cracks
+5°/0.3 (EM)	1	400	fairly long pieces with “ripples” and many side cracks
+5°/0.3 (EM)	2	400	fairly long pieces with “ripples” and many side cracks
+5°/0.5 (EM)	1	ambient	longer pieces with some cracks
+5°/0.5 (EM)	2	ambient	longer pieces with some cracks
+5°/0.5 (EM)	1	400	fairly long pieces with “ripples”, almost no side cracks
+5°/0.5 (EM)	2	400	fairly long pieces with “ripples”, almost no side cracks
+5°/1 (EM)	1	ambient	small pieces (longer than peeling)
+5°/1 (EM)	10	ambient	even smaller pieces than smaller speed
+5°/1 (EM)	1	400	longer than ambient (one fairly long piece but with many cracks on it)
+5°/1 (EM)	10	400	small pieces

Only powder-like chip forms are obtained for $\lambda > 1$, at any pre-heating temperature or deformation rate. Application of hydrostatic pressure is absolutely critical in getting continuous forms in brass 360. Continuous strips formed without pre-heating (Figure 7.5) show indications of severe plastic deformation with significant hardening compared to the bulk. Hardness of the strips are 211 ± 9 and 225 ± 10 HV

for $\lambda = 0.5$ and 0.3 , respectively. Hardness drops sharply to 114 ± 5 HV if the $\lambda = 0.5$ strip is produced at $T_0 = 400$ °C. . The dendritic structures in the as-cast bulk are elongated and sheared at $\lambda = 0.5$, where direction of elongation is parallel to the shear plane. Pores are elongated and smeared in the same direction. The dendritic structure appears to be broken and transformed into the severely deformed matrix at $\lambda = 0.3$ (Fig. 7.5). Pores are again elongated and smeared, but parallel to the chip flow direction at this condition. The gray colored lead pockets can be seen inside some of the pores ($\lambda = 0.3$, high magnification image). This is the solid pure lead that precipitated during the end of the solidification.

For both λ , pores do not interact with each other and do not form cracks when the sufficient hydrostatic pressure is applied. The matrix for both λ values is expected to recrystallize into a very fine grain structure after an annealing treatment.

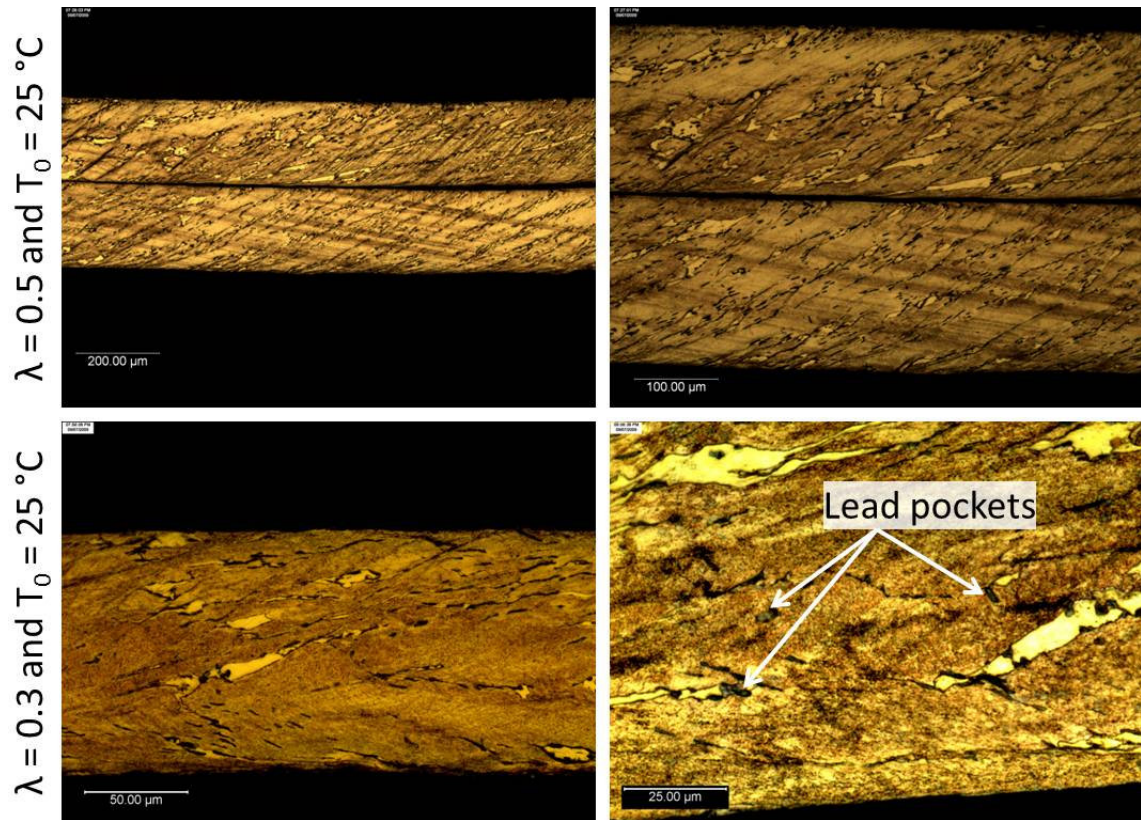


Figure 7-5 Optical micrographs of HP EM brass 360 strips without pre-heating. $\lambda = 0.5$ and $\lambda = 0.3$ are compared. Uniform strips can be obtained with severe plastic deformation and possible ultrafine grain formation in matrix. Lead pockets and other imperfections do not act as crack nucleation sites and do not cause segmentation once the sufficient hydrostatic pressure is applied. $t_0 = 0.3$ mm and $V = 1\text{m/s}$.

7.3 Cast Magnesium AZ31

Magnesium AZ31 is studied extensively in Chapter 6 due to its susceptibility to shear banding. The material studied was in wrought form (tooling plate) with fine grain size. Cast magnesium AZ31 has very large grain size compared to the wrought and it contains porosity. Therefore, segmentation due to the porosity and casting defects becomes a possibility in addition to the shear banding. In order to test the feasibility of EM of limited workability alloys, it is important to study the cast form of an alloy that already has limited workability in wrought form. Furthermore, due to the bigger grain size the cast alloy, deformation mechanisms may change at a

given thermo-mechanical condition, which has a significant effect on the DRX and shear banding behavior.

A pie shape sample is cut from the cross-section of the direct-chill (DC) cast ingot of AZ31 (Figure 7.6). Grain size is on the order of millimeters. Microstructure shows dendritic solidification with porosity and segregation. Pores are trapped near dendrite arms during the solidification (Figure 7.6 right). Grain size is homogeneous through the entire radius of the ingot as a result of the DC casting method (Figure 7.6 left).

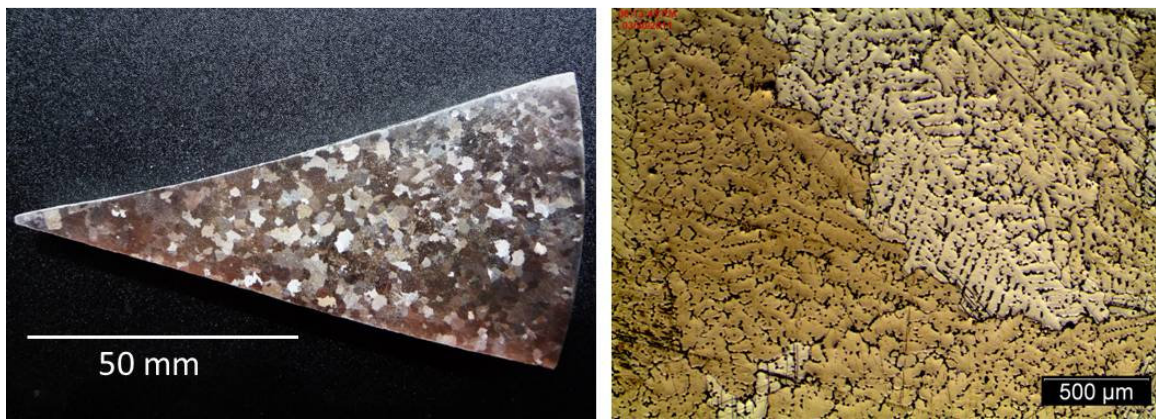


Figure 7-6 Macrostructure (left) and microstructure (right) of the DC cast magnesium AZ31.

Strips with good quality are obtained with EM by fixing $\lambda < 0.9$ (Figure 7.7). Serrated and segmented strips are observed when $\lambda > 0.9$. Initial chip thickness was 0.25 mm and realizing higher chip thicknesses was more difficult with the cast material. Pre-heating was almost always needed for $t_0 > 0.25$ mm. Similar to the other cast materials, side cracking was again a problem and application of side walls resulted in crack free strips.

Microstructure of the strips is shown in Figure 7.7. Room temperature EM with $\lambda = 0.7$, resulted in inhomogeneous microstructure which is mixture of heavily-etched

(dark) and lightly-etched (bright) regions (Figure 7.7 a-c). Heavily-etched regions are uniformly DRX to a very fine ($1 - 2 \mu\text{m}$) grain size (Figure 7.7 b&c). Lightly-etched regions are also believed to be DRX to ultrafine grains (not resolvable under optical microscopy – Fig. 7.7c), although the fraction of DRX should be small in these regions resulting in bright appearance under optical microscopy. Twinning is not observed in lightly-etched regions, given the temperature rise at the high deformation rate ($V = 1 \text{ m/s}$). Therefore, a dislocation slip based deformation mechanism is expected in both regions. Then, the difference between etching responses can be explained by the local strain differences. The lightly-etched regions are believed to be deformed to a lower strain value, causing lesser dislocation and DRX activity.

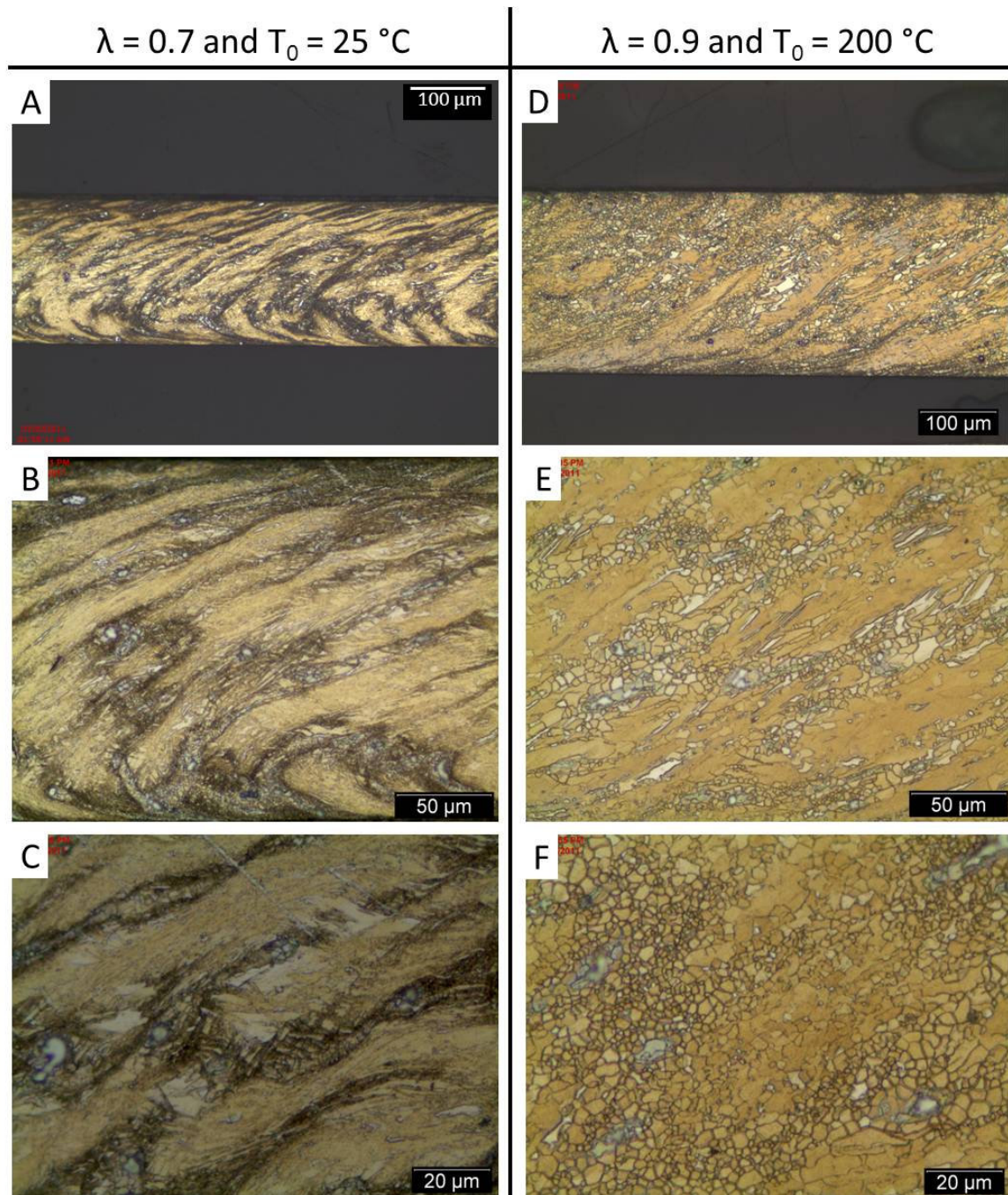


Figure 7-7 Optical microstructures of cast AZ31 processed at $\lambda = 0.7$, RT (a-b) and $\lambda = 0.9$, $T_0 = 200\text{ }^{\circ}\text{C}$ (d-f). Non-uniform deformation is evident with DRX.

The reason of the non-uniformity is believed to be to the dendritic structure of the as-cast material. Each small dendrite arm (Figure 7.6 right) might have been

deformed to a different extent, resulting in a mixture of heavily-etched – higher strain –, and lightly-etched – lower strain – regions.

Inhomogeneous microstructures are persistent, even when the chip thickness ratio and pre-heating temperature are increased simultaneously (figure 7.7 d-f). Heavily-etched regions can be distinguished from lightly-etched regions (Figure 7.7 d&e), with a finer DRXed grain size in the heavily-etched compared to the lightly-etched (Figure 7.7f). Higher strain and corresponding dislocation activity may have caused more nucleation and less growth in the former, whereas lower rate of nucleation is expected in the latter. Growth rate will be high in both cases due to the high deformation temperature. Therefore, complete recrystallization is expected in both regions. The reason of light-etching behavior is believed to be the bigger DRXed grain size.

Therefore, the application of hydrostatic pressure was again critical in deformation of cast AZ31. Continuous forms are obtained devoid of shear bands and cracks. The porosity in the as-cast material is not observed in the final microstructures, suggesting densification through combined effects of hydrostatic pressure and temperature. The non-uniformity of the final microstructures can be overcome by a homogenization treatment before EM. Also, dendrite arm spacing can be decreased by modifying the solidification parameters.

In order to understand the DRX event in cast magnesium, low temperature strips ($T \approx 160\text{ }^{\circ}\text{C}$) are obtained by decreasing the deformation rate to 0.15 m/s. Microstructure again consists of heavily-etched and lightly-etched regions (Figure 7.8). Strain is localized to narrow shear bands in heavily-etched regions. TEM microstructures show clear dynamic recrystallization with very fine grain size $\sim 100\text{ nm}$.

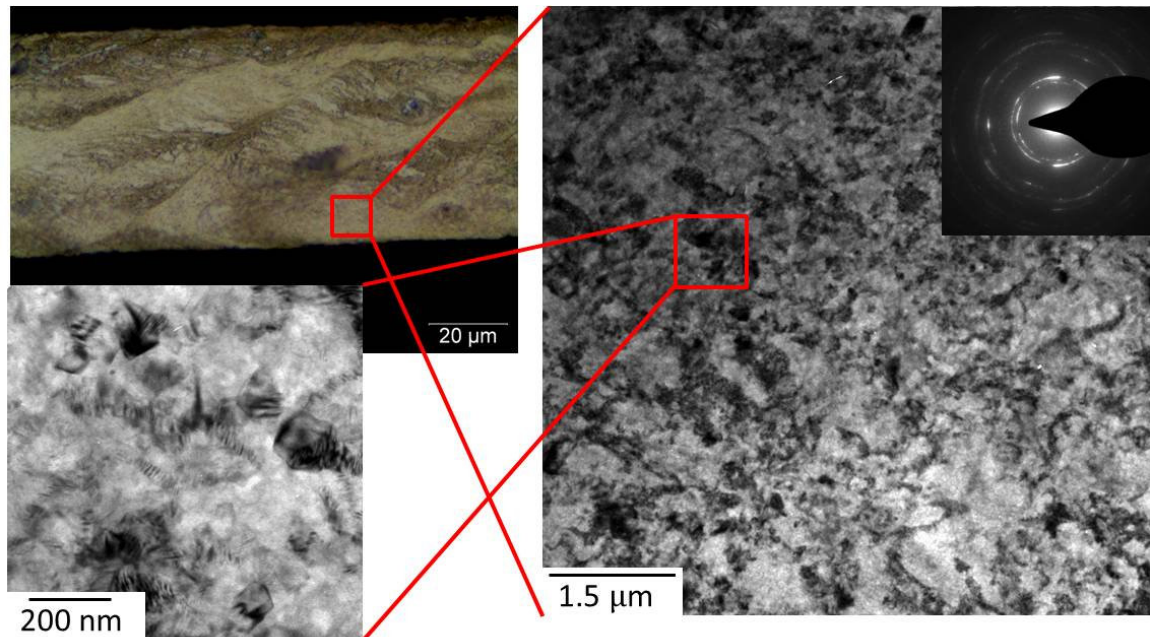


Figure 7-8 Optical and bright-field TEM microstructures of low temperature EM of cast AZ31. Dynamic recrystallization is evident with very fine grain size ~ 100 nm.

These TEM microstructure are very similar to those obtained in strips that are produced with low temperature EM of wrought magnesium (Fig. 6.13). It is possible to suggest that the DRX phenomena occurs even in bigger dendrites in the cast material, as well as the fine wrought grains. Of course, the critical deformation temperature needs to be exceeded for increased dislocation activity, thus DRX. Another important factor is the application of hydrostatic pressure when processing both the cast and wrought materials. This may trigger enhanced dislocation activity and recrystallization.

CHAPTER 8. IMPLICATIONS FOR DEFORMATION PROCESSING AND FUTURE WORK

8.1 Summary of Sheet Making Capabilities of EM and Implications for Deformation Processing

The strip production by extrusion machining is shown to be a controlled deformation process with attributes suitable for enhancing workability of alloys and for effecting SPD, while creating strip forms directly from bulk material. The controllable process parameters are hydrostatic pressure, temperature, strain and strain rate levels and deformation paths. Confinement of the intense deformation field, a key feature of the process, enables highly efficient deformation processing of alloys without pre-heating. These attributes are demonstrated in creation of sheet forms of wrought Mg AZ31B alloy, sintered copper, cast brass 360 alloy and cast Mg AZ31 alloy, all being commercially significant but noted for their poor workability. By in situ control of hydrostatic pressure and temperature in the deformation zone, segmentation in chip formation is suppressed enabling sheet to be created with fine-grained and UFG microstructures. Guidelines in terms of deformation and process parameters for realizing specific sheet properties are presented. EM appears to be quite scalable for commercial production of sheet and foil. Lab-scale strips of up to 50 mm width and 2 mm thickness are realized with a small capacity lathe.

In the first part of the thesis (Chapter 6), wrought Mg AZ31 was processed to show the adiabatic shear banding suppression capabilities of EM. ASB behavior is explained as a function of deformation and dynamic recrystallization mechanisms which are highly dependent on the processing parameters. Magnesium AZ31 switches its deformation mechanism and shows a sharp increase in plasticity with

the temperature. The switch in the deformation mechanism is noted by the sharp increase in the strain-rate sensitivity and increased dislocation activity. Shear banding becomes impossible when the switch in the deformation mechanism happens with the realization of the critical temperature. The change in strain rate within the range studied in this work ($10^3 - 10^5/s$) has minimal effects on the deformation mechanisms and shear banding behavior at a constant deformation temperature and an effective strain of 1.15. Therefore, the shear banding behavior is essentially a function of the deformation temperature. The shear banding models are modified for temperature effects. Modified model can predict the onset of the uniform flow behavior by the proper selection of flow hardening/flow softening rate. A rate of $n'_\tau = -0.30$ is found to be a successful input to the model for prediction of the critical deformation temperature above which uniform flow is possible.

The shear banding behavior, deformation mechanisms and the microstructure evolution are explained in three temperature regimes. At low temperatures, $80 < T < 150^\circ\text{C}$, twinning and basal slip are the dominant deformation mechanism. Twinning induced dynamic recrystallization (TDRX) and rotational dynamic recrystallization (RDRX) happens only to a limited extend because of the limited dislocation activity. Microstructure becomes highly non-uniform on the local level and shear localization is possible to these DRX regions followed by the catastrophic shear banding due to the lower values of strain-rate sensitivity. This results in the non-uniform microstructure on the thickness scale, which is a mixture of DRX shear bands and a lesser strained matrix region consisting of original wrought grains. Cracks are usually associated with shear bands causing the segmentation and disintegration of the strip. At high temperatures, $230 < T < 300^\circ\text{C}$, non-basal slip and cross-slip are dominant. CDRX happens everywhere in the microstructure due to the increased cross-slip activity. Shear localization to these regions is not possible due to the local uniformity of the microstructure. Even in the case of shear localization, further catastrophic shear banding is prevented due to the high values of strain-rate sensitivity. CDRX event continues to happen uniformly throughout the

microstructure, resulting in strips devoid of shear bands and cracks. At intermediate temperatures, $150 < T < 230^{\circ}\text{C}$, the deformation and shear banding behavior is a mixture of low and high temperature regimes. However, susceptibility to shear banding decreases with increasing temperature since microstructure becomes more homogeneous with the fraction of CDRX regions increasing. Shear banding does not happen when CDRX happens uniformly throughout the microstructure and the strain-rate sensitivity increases.

In both intermediate and high temperature regimes, very fine grain sizes, 150 nm – 3 μm are obtained together with non-basal deformation textures, which are both essential in further processing of the magnesium strip.

In the second part of the thesis (Chapter 7), cast brass 360, cast Mg AZ31 and sintered pure copper are processed in order to further test the capabilities of EM. These materials are selected for their extrinsic susceptibility to segmentation and cracking due to the defects such as impurities, imperfections, weak second phases, porosity and large grain sizes. Continuous strips with good quality are possible with combinations of hydrostatic pressure and deformation temperature. Hydrostatic pressure becomes critical for getting crack free forms when working with cast materials. Side cracking is usually observed in this class of materials and it can be prevented by implementing side walls in the EM tool.

For the case of sintered pure copper, which has high workability in the wrought form, cracking and segmentation is not a problem even for higher chip thickness ratios up to 1.8. Pre-heating is not necessary for getting continuous forms. Pore closure and densification is also observed at $\lambda = 1.8$ with some residual porosity. Densification increases with decreasing chip thickness ratio resulting in lower residual porosity levels. The residual pores are elongated and smeared parallel to the shear plane in the case of higher chip thickness ratios. They are oriented parallel to the chip flow direction as chip thickness ratio decreases. The uniform grain size in

the as-received sintered material results in uniform flow and uniform, severely deformed microstructures.

Cast brass 360 is more challenging since the wrought form of this alloy -itself- has limited workability. Combinations of high hydrostatic pressures ($\lambda < 0.5$) and high pre-heating temperatures ($T_0 = 400\text{ }^{\circ}\text{C}$) are crucial in preventing cracking which is likely originated from casting porosity and associated lead pockets inside the porosity. Both the dendritic structure and the porosity shears and elongates after room temperature extrusion machining at $\lambda = 0.5$. The dendritic structure is broken and some pore closure is observed at a lower chip thickness ratio of $\lambda = 0.3$.

Cast Mg AZ31 shows the most non-uniform deformation behavior. Cracking and segmentation are not problems for this alloy for $\lambda < 0.9$, once the critical deformation temperature is exceeded ($T > 230\text{ }^{\circ}\text{C}$) as in the case of wrought alloy. Dendrites in the as-cast microstructure are again sheared and elongated after EM. On the local scale, however, each dendritic arm shows a different deformation and DRX behavior. TEM microstructures confirm that DRX indeed happens inside the dendrites above $160\text{ }^{\circ}\text{C}$, in spite of their large size. However, the extent of DRX changes from one dendrite to another resulting in various DRXed grain sizes. The non-uniform microstructures are persistent even at higher deformation temperatures realized by pre-heating. The non-uniformity of the microstructure may adversely affect the mechanical properties. Porosity in the as-cast microstructure is not a problem and very minor residual porosity is observed after EM.

8.2 Feasibility of EM for Sheet Making and Recommendations for Future Work

Extrusion machining has the most cost and energy savings potential if the strips are produced directly from the cast feedstock. Both hot rolling and cold rolling steps are skipped if the feedstock is cast. Wrought feedstock may not provide enough cost advantage since only cold rolling is bypassed in this case. The process itself is quite

scalable if provided with powerful and sturdy machinery. While the strips dimensions can be selected based on the final application, the process is applicable in production of the strips that are < 300 mm in width and < 2 mm in thickness. The process becomes more competitive as the final strip thickness decreases since it is possible to skip many hot and cold rolling steps. It is also possible to reduce the inventory costs and lead times for small batches of production, as EM can produce the exact width without the need for stocking wider sheets and further slitting them. Therefore, cast feedstock need to be studied further. As-cast billets from various techniques should be tested with EM and for a given material system. Cast billets should be selected that result in an optimum strip at low cost. The microstructure of the cast-billet should be refined to the minimum possible scale. A homogenization treatment may be necessary before the EM.

Below table briefly compares the rolling and extrusion machining processes in terms of their capabilities, advantages and disadvantages.

Table 8-1 A comparison of rolling and extrusion machining processes

	Rolling	Extrusion Machining
Deformation Zone	Wide	Narrow (enables local temperature rise)
Effective strain ($\bar{\epsilon}$)	0.2 - 0.5 (per pass)	> 1 (single pass)
Deformation Rate	up to 10 m/s	up to 5 m/s
Texture	Rolling type	Shear type
Microstructure	Fine grain (> 10 μm) after final annealing	Ultrafine to very fine grain (0.2 – 10 μm)
Energy Requirements	Pre-heating, intermediate annealing, friction losses	Specific shear energy, friction losses, limited pre-heating
Advantages	Large capacity, high volume production, capable of large dimensions	Cost, raw material and energy efficient, small batch and flexible production, compact design and less capital requirement
Challenges	Multi-pass, heat losses	Scale-up, strip flatness and camber, initial cast feedstock, tool wear
Disadvantages	Capital intensive, inefficient, higher lead times and inventory needs, material losses	Limited width and thickness

A scale-up in the strip dimensions would be the next logical research step. Strips should be obtained with uniform thickness and width at the increased dimensions. Camber and flatness issues that can arise with scale-up step should be solved. A simple strip tensioning step can be integrated to the EM, as it is shown to reduce

camber and increase the flatness of the strips [20, 21, 105]. Applied strip tension also reduces the cutting forces and increases the tool life.

The tool life was concluded not to be a major issue in peeling even when the starting material was stainless steel [20]. However, it still needs to be characterized for various material systems as it can be a limiting factor in cost efficiency. Processing parameters should be optimized for highest possible production rate at minimal tool wear. Cutting fluids may need to be employed to reduce the tool wear and the cutting forces.

Rake angle is another processing parameter that can be optimized. It was kept constant (and high) in this thesis, for the purpose of imposing large strains in the deformation zone for the highest possible level of microstructure refinement. At a given chip thickness ratio, a more positive rake angle can significantly decrease the cutting forces (up to 40%) and tool wear at a cost of lesser effective strain. Cracking and segmentation tendency reduces with increasing rake angles and higher initial chip thicknesses can be realized. Side cracking also diminishes at positive rake angles.

The effects of deformation rate, temperature and hydrostatic pressure on the segmentation and shear banding behavior are explained parallel to the models, while the effects of increasing chip thickness remain unexplained. An increase in initial chip thickness beyond some critical level results in segmented chips in any given material system. Pre-heating of the workpiece was almost always necessary in order to get continuous strips after increasing the initial chip thickness beyond the critical level. Therefore, the mechanisms of segmentation with higher feed rates should be investigated. Some experiments are proposed in section 6.1.3 for this study.

Last, but not least, the final mechanical properties and formability of the strips should be tested and compared to the standards (ASTM). The scale-up in strip dimensions is needed in order to test mechanical properties and formability on the commercial scale. However, very fine grain sizes together with shear-type deformation textures (non-basal in the case of magnesium) that are possible with EM suggests an increase in the formability compared to the rolled strips.

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Research Interests:

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Funded Research Proposals:

- S. Chandrasekar, K.P. Trumble and M. Efe (Co-PI), "Enhancing Deformation Processing Capability of Structural HCP Metals," 2011-2013, Funded by Army Research Office, Program Manager: S.N. Mathaudhu

Journal Publications:

- M. Efe, H.J. Kim, S. Chandrasekar, K.P. Trumble, "The chemical state and control of oxygen in powder metallurgy tantalum," *Material Science and Engineering A*, 544, May 2012, p1-9.

- M. Efe, W. Moscoso, K.P. Trumble, W.D. Compton, S. Chandrasekar, "Mechanics of Large Strain Extrusion Machining and Application to Deformation Processing of Magnesium Alloys," *Acta Materialia*, 60, March 2012, p2031-2042.

- Y. Guo, M. Efe, W. Moscoso, D. Sagapuram, K. P. Trumble and S. Chandrasekar, "Deformation field in large strain extrusion machining and implications for deformation processing", *Scripta Materialia*, 66-5, March 2012, p235-238.

- D. Sagapuram, M. Efe, W. Moscoso, S. Chandrasekar, K.P. Trumble, "Deformation Temperature Effects on Microstructure and Texture Evolution in High Strain Rate Extrusion-Machining of Mg-AZ31B," *Material Science Forum*, Vol. 702-703, 2011, p52.

- O. Atwani, D.V. Quach, M. Efe, P.R. Cantwell, B. Heim, B. Schultz, E.A. Stach, J.R. Groza, J.P. Allain, "Multimodal grain size distribution and high hardness in fine grained tungsten fabricated by spark plasma sintering," *Materials Science and Engineering: A*, 528, July 2011, p5670-5677.

Patent Applications:

- S. Chandrasekar, K. Trumble, W. Moscoso, M. Efe, D. Sagapuram, C.J. Saldana, J.B. Mann, W.D. Compton, "Large-Strain Extrusion Machining Processes and Bulk Forms Produced Therefrom", US Patent Application # 12/020498, Filed on: 1/6/2012

Conference Presentations & Publications:

- M. Efe, D. Sagapuram, W. Moscoso, S. Chandrasekar, K.P. Trumble "Understanding the Ultrafine Grain Formation and Recrystallization Mechanisms in Magnesium through Extrusion-Machining," Presentation at TMS 2012, Orlando, FL, March 2012, *TMS Young Scientist Silver Medal*
- M. Efe, W. Moscoso, W.D. Compton, K.P. Trumble, S. Chandrasekar, "Direct Production of Sheet from Alloys of Limited Workability Using Machining-Based Processes," Presentation at TMS 2011, San Diego, CA, March 2011
- M. Efe, W. Moscoso, S. Chandrasekar, K.P. Trumble, "Production of Magnesium AZ31B Foils by a Machining-Based Processes", Presentation at MS&T 2009, Pittsburgh, PA, October 2009
- M. Efe, H.J. Kim, S. Chandrasekar, K.P. Trumble, "Hot-Pressing of Ultrafine Grained Tantalum Powders Produced from Machining Chips", in: Proceedings of the 2009 International Conference on Powder Metallurgy & Particulate Materials, MPIF, Princeton, NJ, p8-87:8-96, June 2009
- M. Efe, H.J. Kim, W. Moscoso, W.D. Compton, S. Chandrasekar, K.P. Trumble, "Heavily Deformed Tantalum Plates by Linear Extrusion Machining", Poster Presentation at TMS Annual Meeting, New Orleans, LA, March 2008

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