THE PORTEVIN-LE CHATELIER EFFECT
AND SHEAR BAND FORMATION
IN AA5754 ALUMINUM ALLOY
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TITLE: The Portevin-Le Chatelier Effect and Shear Band Formation in AA 5754 Aluminum Alloy


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The use of AA 5754 Al-Mg alloy for automotive applications is limited by its rapid shear failure process, due to shear banding. This failure mechanism is further complicated by the presence of inhomogeneous plastic deformation, so-called Portevin-Le Chatelier (PLC) effect, during deformation. Therefore, the purpose of this study was primarily to investigate the impact of Portevin-Le Chatelier (PLC) banding towards shear banding in this commercial alloy. The second objective was to study the PLC banding as a function of prior deformation under positive strain rate sensitivity condition.

The experimental work involved pre-straining experiments coupled with a non-contact strain measurement technique. Pre-straining experiments were carried out by deforming the sample at 223 K, at which the PLC effect is significantly suppressed, up to a prescribed amount of true strain prior to room temperature testing. A non-contact strain measurement technique, based on digital image correlation (DIC), was utilized in order to observe PLC band behavior during tensile tests at room temperature and subsequently to measure the amount of plastic strain carried within the band.

The results showed the appearance of random nucleation deformation bands, associated with type B PLC banding, with short distance propagation during constant strain rate tensile test at room temperature. A change in the nature of PLC banding, marked by distinct band propagation, was observed once a critical amount of pre-strain is given. However, there is no evidence of a relationship between two existing phenomena, PLC banding and shear banding, in this alloy.
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CHAPTER 1

INTRODUCTION

In the last decade, the use of aluminum alloys as automotive materials has increased significantly due to the demand for higher fuel efficiency and lower emission vehicles. Aluminum alloys offer not only high strength to weight ratio but also good corrosion resistance, which enable the mass of vehicles to be economically reduced without sacrificing their structural performance. In general, there are two main series of aluminum alloys which have been considered for automotive industrial applications. The 5000 series solution hardened alloys are used for unexposed structural applications while the 6000 series precipitation hardened alloys are preferred for skin panels. The preferred alloys in North American cars are AA5754 and AA6111 (Miller et al., 2000). Therefore, the focus of this study is on the 5754 Al-Mg alloy, which offers great potential applications, yet possesses many interesting phenomena that have not been fully understood up to this point.

The 5000 series alloys exhibit repeated discontinuous yielding, the so-called Portevin-Le Chatelier (PLC) effect, over a specific regime of strain rate and temperature. It is generally accepted that the PLC effect is attributed to dynamic strain aging (DSA), in which solute atoms of the alloy interact with mobile dislocations during deformation processes. This PLC effect leads to a reduction in ductility and leaves macroscopic surface band markings, and hence limits the potential applications of the alloys.
Tremendous research has been performed to study PLC effect experimentally on both high purity grade (Nortmann and Schwink, 1997; Klose et al., 2004) and commercial grade (Wen and Morris, 2003; Shabadi et al., 2004) alloys. Different types of PLC band, namely type A, B, and C, have been classified based on displacement controlled uniaxial tensile testing. Finite element analysis has been successfully utilized to simulate various types of PLC bands, corresponding to those which have been observed experimentally (Zhang et al., 2001; Lasko et al., 2005).

Moreover, it has been well known that Al-Mg alloys often fail by plastic flow localization into narrow bands of intense shear (Korbel et al., 1986; Bird et al., 1987). A recent study on 5754 Al-Mg alloy (Spencer et al., 2002) showed that it does not exhibit the classic behavior of ductile fracture, which involves the nucleation of voids followed by voids growth and ultimate linkage. Instead, the failure mode of this alloy was shear banding with little evidence of damage prior to final stage of fracture. This finding suggests that shear band formation accelerates ductile fracture during the plastic deformation process.

The two different phenomena, PLC banding and shear banding, occurring in 5754 Al-Mg alloy must be carefully investigated as they limit the applications of the alloy in automotive industries. Therefore, one of the objectives of this work is to study the interactions between these two phenomena, in order to be able to optimize the parameters of production processes of the alloy. The second objective is to characterize PLC banding as a function of prior deformation under positive strain rate sensitivity conditions. As will be shown later in this work, the PLC effect is significantly suppressed at 223 K.
Therefore, the study of the impact of PLC effect toward the final shear localization process is carried out by deforming the samples up to different value of true strain at 223 K. The samples were then reloaded at room temperature.

Chapter 2 presents a literature survey of various experimental studies, simulations, and theories that have been developed on describing PLC effects and shear banding. Chapter 3 discusses details of the experimental setup, which combines uniaxial tensile tests with digital image correlation (DIC) techniques, followed by the experimental results in Chapter 4. Chapter 5 presents a discussion of the experimental results obtained in this study. Finally, the highlights of the discussions based on the experimental results corresponding to the initial research objectives will be summarized in Chapter 6.
CHAPTER 2

LITERATURE REVIEW

Many experimental observations and modeling studies have been performed to understand and explain the Portevin-Le Chatelier (PLC) effect in metallic alloys. Experimentally, the PLC effect has been studied by deforming materials mostly in tension (Chihab et al., 1987; Hähner et al., 2002; Shabadi et al., 2004; Zhang et al., 2005). Correspondingly, numerical modeling (McCormick and Ling, 1995; Hähner and Rizzi, 2003) and finite element simulations (Zhang et al., 2001; Kok et al., 2002; Lasko et al., 2005) have been performed in order to gain better understanding of this effect.

In addition to the PLC effect studies, the formation and development of shear bands in metallic alloys have also been extensively investigated by deforming the materials in both tension and rolling (Brown, 1972; Korbel et al., 1986; Bird et al., 1987). To further comprehend the background, this chapter will outline salient experimental and modeling works on the PLC effect and shear banding in metallic alloys, particularly in polycrystalline Al-Mg alloys.

2.1. Theory of Dynamic Strain Aging

Plastic deformation in crystalline solids, which occurs due to the motion of line defects, so-called dislocations, is inherently non uniform at the microscopic scale. It is therefore common for plastic flow to become unstable and be localized macroscopically
during deformation to finite plastic strain. The PLC effect, which manifests itself as a serrated stress-strain curve, is one type of plastic deformation instability initially associated with negative strain rate sensitivity by Penning (1972). Strain rate sensitivity of flow stress plays an important role in determining macroscopic behavior of metallic alloys. As shown by Penning (1972), inhomogeneous deformation, as observed in the PLC effect, can be explained by assuming that strain rate sensitivity is negative in a finite interval of strain rate.

Penning's phenomenological theory was depicted as an N-shaped curve, which relates the flow stress to strain rate, as shown in Figure 2.1. The theory stated that the negative strain rate sensitivity interval, which is the section of negative slope, is inaccessible. Therefore, during the loading process the plastic strain rate jumps instantaneously from point c to d, but plummets from point f to a during the unloading process. Such a drastic strain rate change is possible because two (or more) solutions exist for $\dot{\varepsilon}$ at the same stress level.

Microscopically, it has been generally accepted that dynamic strain aging (DSA), in which solute atoms interact with mobile dislocations during deformation processes, is the physical origin of the PLC effect. Solute mobility introduces a negative contribution to the total strain rate sensitivity of flow stress and this contribution increases with increasing strain. Once the total strain rate sensitivity becomes negative, plastic flow becomes unstable and the characteristics of the PLC effect start to appear.
Figure 2.1 N-shaped curve describing the flow stress as a function of plastic strain rate $\dot{\varepsilon}^p$, with associated stress-strain response. Numerical values are representative of an Al-Mg alloy at room temperature (Kok et al., 2002).

The dynamic strain aging phenomenon was firstly explained by Cottrell (1953) based on the concept of dislocations moving with a steady velocity and dragging a solute atmosphere at the same velocity. In interstitial alloys, this concept was supported by the fact that the solute atoms could diffuse fast enough to allow aging during deformation. Later on, Cottrell claimed that production of vacancies through plastic deformation could further enhance solute atoms diffusion in substitutional alloys to allow dynamic aging. This idea was then used to clarify the existence of uniform deformation, so-called critical strain, prior to serrated yielding. Generally, in substitutional alloys, the critical strain increases with increasing strain rate and decreasing temperature.

About two decades later, McCormick (1972) pointed out that the movement of dislocations is actually discontinuous, in jerky flow, instead of steady, as was claimed
previously by Cottrell. His theory stated that mobile dislocations are arrested at localized obstacles, i.e. forest dislocations, throughout their movement. Consequently, the interaction between mobile dislocations and diffusing solute atoms can mainly occur during specific intervals, i.e. the periods of time spent at obstacles, so-called waiting time. For such a process, the average dislocation velocity $\bar{v}$, can be expressed in terms of a waiting time $t_w$, and a time of flight through the lattice to the next obstacle $t_f$.

$$\bar{v} = \frac{L}{t_w + t_f} \quad \ldots (2.1)$$

where $L$ is the average distance between arresting obstacles. In most cases, waiting time is much larger than time of flight, therefore the average dislocation velocity is determined primarily by the waiting time. The aging process occurred if the waiting time of the dislocation at an obstacle is sufficiently long. There is no need for long range diffusion of substitutional solute atoms, hence only local diffusion of solute is required for dynamic strain aging to take place.

This theory has been further developed by Van den Beukel (1975), who calculated the dependence of flow stress on strain rate, temperature, and strain in the presence of dynamic strain aging, resulting in a formulation corresponding to the change in strain rate sensitivity. It was noticed that local solute composition at arrested dislocations is not constant when dynamic aging takes place, but is instead a function of waiting time and solute diffusion coefficient. The negative strain rate dependence of flow stress was possibly due to the diffusion of solute atoms into dislocations temporarily arrested at obstacles on their slip path.
2.2. Spatial Characteristic of Portevin-Le Chatelier (PLC) Band

In general, three types of PLC band can be classified based on the type of serration that appeared in the stress-strain curve of polycrystalline materials during constant strain rate tensile tests, as shown in Figure 2.2. The appearance of the serrations changes from type A to type B and to type C, either with decreasing strain rate or increasing temperature. Type A serration is associated with repetitive continuous propagation of deformation bands along the gauge length of the sample (similar to Lüders bands) and is often nucleated at one end of gauge length. Type B serration corresponds to a hopping propagation of localized bands in the axial tensile direction of the specimen. Type C serration is characterized by random nucleation of localized deformation bands accompanied with a large stress drop in the stress-strain curve. These various surface-related characteristics of PLC bands require a development of techniques, which enable detailed observation of the sample’s surface during the deformation process.
Figure 2.2 Schematic stress-strain curves showing types A, B and C serrations in a Cu-In alloy (Brindley and Worthington, 1970).

Realizing the importance of aforementioned issue, the in-situ surface observation technique during deformation process was vigorously developed, which enabled detailed investigations of spatial characteristics, namely band strain, band width, and band velocity, of PLC bands. In early work of Chihab et al. (1987), the characteristics of PLC bands of Al-5 at.% Mg were critically examined using a TV camera on a flat tensile sample. Type A bands occurred at strain rates higher than $5 \times 10^{-3} \text{ s}^{-1}$, type B bands in the strain rate interval of $10^{-4} \text{ s}^{-1}$ to $10^{-3} \text{ s}^{-1}$, and finally type C bands at strain rate lower than
$5 \times 10^{-5}$ s$^{-1}$. Band width and serration amplitude were found to decrease with increasing strain rate, as the type of the band changed from C to B and to A, as shown in Figure 2.3. Typical band width of 1.5 mm was observed when type C serrations occurred, while type B bands had typical band width of 0.6 mm. Overall, band width shows decreasing trend with increasing strain rate. This TV camera technique, however, was lacking in capability of measuring local plastic strain within the PLC band, so-called band strain.

![Graph showing strain rate dependence of PLC band width and serration amplitude](image)

**Figure 2.3** Strain rate dependence of PLC band with (solid circles) and of serration amplitude (open circles) (Chihab *et al.*, 1987).

The evolution of PLC bands in Cu-15 at.% Al was investigated in details by using a laser scanning extensometer technique on a flat tensile sample (Hähner *et al.*, 2002). In all cases, during tensile tests below 425 K at a given strain rate, whose interval was between $6.67 \times 10^{-6}$ s$^{-1}$ and $6.67 \times 10^{-3}$ s$^{-1}$, there was a direct transition from PLC type A
to type C with increasing strain. As shown in Figure 2.4, band width ($w_b$), band strain ($\Delta \varepsilon_b$) and band velocity ($c_b$) exhibited an increasing trend with increasing strain rate at 373 K. This observation is not in agreement with the results of Chihab et al. (1987) who reported a decreasing trend with increasing strain rate in Al-Mg alloy. It should be noted, however, that most of Chihab’s measurement refer to PLC type B.

![Double logarithmic plots of band strain, band width, and band velocity as functions of applied strain rate at 373 K (Hähner et al., 2002).](image)

Figure 2.4 Double logarithmic plots of band strain, band width, and band velocity as functions of applied strain rate at 373 K (Hähner et al., 2002).

Following the work of Hähner et al., Klose et al. (2004) examined the band strain and accompanying stress drops of type B serration in the same material. It was found that band strain, whose maximum value varies between 0.2% and 0.35%, increased with increasing strain. The magnitude of stress drop increased up to a total strain of 10%, yet
decreased at larger strain. This reflects an increase in dynamic strain aging effectiveness with increasing dislocation density (Springer et al., 1998). At larger strain, however, solute exhaustion (Springer et al., 1998) or decrease in dislocation path length (Chmelik et al., 2002) or combination of both would reduce the effectiveness of dynamic strain aging process.

Shabadi et al. (2004) utilized a laser speckle technique to perform in situ observation of PLC bands in commercial grade AA5082 Al-Mg alloy. At strain rate of $10^{-3} \text{ s}^{-1}$, type B serration was observed for which the magnitude of the stress drop gradually increased and then saturated with strain. Note that the result contradicted Klose’s observations on type B serration in Cu-15 at.% Al. Multiple deformation bands were observed at random locations along the gauge length at a fixed angle ranging from $55^0$ to $65^0$ with respect to tensile axis. Band width was found to increase with increasing strain, whereas band velocity decreases.

Recently, a non-contact strain measurement, so-called digital image correlation (DIC) technique, has been used to study the kinematics of PLC deformation bands. Jiang et al. (2005) investigated the spatial features of PLC bands in commercial AA2017 Al-Cu alloy. Type A, B and C bands were observed in the range of strain rate between $2.9 \times 10^{-5} \text{ s}^{-1}$ and $2.9 \times 10^{-3} \text{ s}^{-1}$ at room temperature. The velocity of type A and type B bands decreased gradually with increasing strain, while type C bands just displayed a stochastic nucleation nature with no propagation. It was also observed that propagation velocity slowed down distinctly as characteristic of PLC band changed from type A to type B.
An individual type B PLC band in another commercial Al-Mg alloy, i.e. AA5052, was studied with the DIC technique (Tong et al., 2005). At the temporal resolution of 0.2 ms and the spatial resolution of about 20 µm, the type B PLC band was found to be purely stationary at the macroscopic level as no plastic deformation front was detected to propagate either along the band’s length direction or along the band’s width direction (the tensile loading direction). It was also found that the strain distribution across the PLC band width showed a characteristic bell-like strain profile with a peak strain.

With the same technique, Kang et al. (2006) investigated the behavior of type B PLC bands in AA5754 Al-Mg alloy during uniaxial tension test. No PLC band propagation was observed during the load drop phase in the stress-strain curve. During the increasing load phase, however, PLC bands appeared and swept across the sample along the tensile axis. Moreover, the strain increment associated with PLC bands, so-called PLC band strain, increased with increasing strain.

During tensile tests with constant strain rate, the band strain increases, while the band velocity decreases, with increasing strain. However, the dependence of band width and band velocity on strain rate is still inconclusive due to contradictory results available in the literature.

2.3. Numerical and Finite Element Modeling of PLC Effect

An early mathematical model describing the PLC effect under constant stress rate was formulated by Kubin and Estrin (1985) resulting in an exact analytical description of
discontinuous deformation behavior. The formulation was based on Penning’s original model:

\[ \sigma = h \varepsilon + F(\dot{\varepsilon}) \]  \hspace{1cm} (2.2)

where \( h \) is the work hardening rate, considered constant and positive, and \( F(\dot{\varepsilon}) \) is a strain rate dependent function, whose behavior gives rise to a strain rate dependence of \( \sigma \). Simplified Bailey-Orowan formulation was used in order to take into account the effect of static recovery. It was shown that band width and band strain were controlled by intrinsic material properties, i.e. work hardening rate, recovery rate, strain-rate sensitivity function \( F(\dot{\varepsilon}) \), and by stress rate and independent of specimen length.

Three years later, McCormick (1988) attempted to develop a constitutive equation for plastic flow localization due to dynamic strain aging. The flow stress equation may be described by:

\[ \sigma = \sigma_i + S_i \left\{ \ln \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right) + H'_c C_s \right\} \]  \hspace{1cm} (2.3)

where \( \sigma_i \) is the internal stress, \( S_i \) the instantaneous strain rate sensitivity, and \( H'_c \) was expressed as:

\[ H'_c = \left( \frac{1}{kT} \right) \left( \frac{\partial H}{\partial C_s} \right) \]  \hspace{1cm} (2.4)

where \( H \) is the activation enthalpy for dislocation motion. An inherent feature in this model is a state parameter, \( C_s \), which depends on solute concentration at arrested mobile dislocations. In dynamic strain aging regime, therefore, both \( \sigma \) and \( H \) are affected by a time dependent solute composition at mobile dislocations, \( C_s \).
Based on this constitutive equation, McCormick and Ling (1995) performed numerical simulations which include a Bridgman-type factor to provide an interaction between adjacent straining elements. A geometrical inhomogeneity was introduced which gives rise to strain rate variation over the length of a cylindrical specimen. They successfully simulated type A serrations at the strain rate of $2 \times 10^{-3} \text{s}^{-1}$. Band strain was found to increase almost linearly with increasing strain, while band width initially decreased rapidly with increasing strain and then saturated and its minimum value is approximately equal to the specimen diameter.

Following their work, Zhang et al. (2001) carried out three dimensional finite element simulation of PLC effect in Al-Mg-Si alloy for tensile deformation of flat and round specimens based on a similar constitutive equation. A random distribution of initial external stress, ranging from 0 to 10 MPa, was introduced across the entire specimen in order to trigger strain localization. Planar deformation bands oriented at $35^0 - 37.5^0$ with respect to the tensile axis were observed in flat specimens and the corresponding angle is independent of sample thickness. Band width was found to be approximately equal to sample thickness. Double conical shapes of localized deformation bands which propagate along the round sample were observed.

Hähner et al. (2002) formulated PLC effect based on McCormick’s concept that aging occurs when mobile dislocations are temporarily arrested at localized obstacles, which can be overcome by thermal activation. The idea was expressed following an Arrhenius law for the plastic strain rate, $\dot{\varepsilon}^p$, as follows:
\[ \dot{\varepsilon}^p = \nu \Omega \exp \left[ -\frac{\Delta G + G_0}{kT} + \frac{\sigma_{\text{eff}}}{S_0} \right] \] (2.5)

\( \nu \) is the attempt frequency of thermal activation, \( \Omega \) is the elementary strain accomplished when all mobile dislocations have been activated once, \( G_0 \) is the basic activation enthalpy in the absence of dynamic strain aging, \( \Delta G \) is an additional contribution related to dynamic strain aging. Thermal activation is facilitated by effective stress, \( \sigma_{\text{eff}} \) which drives dislocation motion,

\[ \sigma_{\text{eff}} = \sigma_{\text{ext}} - \sigma_{\text{int}} \] (2.6)

resulting from the externally applied stress \( \sigma_{\text{ext}} \) (flow stress) which is diminished by the internal stress \( \sigma_{\text{int}} \) (athermal back stress). The essence of this model is that additional activation enthalpy \( \Delta G \) related to the kinetics of dynamic strain aging is taken as a dynamic internal variable of the model. The evolution rule for \( \Delta G \) is assumed to be:

\[ \Delta \tilde{G} = \eta (\Delta G_\infty - \Delta G) - \frac{\dot{\varepsilon}}{\Omega} \Delta G \] (2.7)

The first term on right hand side of equation describes an aging process linked to dislocation pinning by solute atoms, while the second one describes dislocation unpinning by thermal activation. This model was considered successful in predicting kinematical characteristics of PLC band, namely band width, band velocity, and band strain. Band width increases linearly with increasing cross-head velocity, whereas band velocity depends non-linearly on applied cross-head velocity. Band velocity increases as the material hardens at a faster rate, while band strain decreases. It suggests that strain hardening limits the amount of plastic strain carried by the plastic strain.
Based on the work of Hähner (2002), Lasko et al. (2005) performed three-dimensional finite element simulations of PLC effect on flat specimen under different applied strain rates. A 5% initial perturbation of back stress was introduced in order to trigger strain localization. Band width was found to depend on the mesh used in simulations; the finer the mesh, the thinner the band width. The band propagated at an angle between $45^0$ and $53^0$ with respect to tensile axis and its velocity increases linearly with applied strain rate.

A common feature of aforementioned models is the necessity of an ‘artificial’ inhomogeneous initial state to simulate propagating deformation bands. McCormick and Ling (1995) used an initial geometric inhomogeneity, Zhang et al. (2001) introduced an initial stress inhomogeneity, and Lasko et al. (2005) used an initial perturbation of back stress. This limitation can be overcome by making use of crystal plasticity, in which the spatial variation in the initial grain orientations (texture) induces a non-uniform stress state. Kok et al. (2002) have developed a crystal plasticity model, embedded in a finite element framework to describe the PLC effect in tension test. The transition from continuous to discontinuous band propagation, which is experimentally observed, can be simulated as the applied strain rate was decreased.

2.4. Shear Failure

In general, ductile fracture of commercial metals and alloys involves two basic modes, namely the fibrous mode and the shear mode. Fibrous fracture is characterized by the formation and growth of voids from inclusions or particles resulting in a fracture
surface covered with equiaxed dimples. The shear mode of ductile fracture gives rise to fracture surfaces covered with shallow and strongly-oriented parabolic dimples. Due to its occurrence in the 5754 Al-Mg alloy, the shear mode of ductile fracture will be further discussed in detail.

2.4.1. Experimental Observations of Shear Localization

Some of the earliest systematic studies of shear localization in commercial grade materials were carried out by French and Weinrich (1973, 1975). They tested cylindrical tensile samples of a brass and commercial copper under hydrostatic pressure in the range of 0.1 – 600 MPa. It was found that from 0.1 MPa to 300 - 350 MPa, the samples failed by the classic cup and cone ductile fracture mechanism. Both a brass and copper samples fractured by a shear process at higher superimposed pressures. Nevertheless, the post-necking behaviour between a brass and copper differed in that copper samples failed by necking down to a point, whereas a brass failed by the void sheet mechanism. It should be noted as well that shear localization did not take place in any case until necking commenced.

Further studies by Weinrich and French (1976) were focused on the same materials in the sheet form instead of cylindrical. It was noticed that even though imposed hydrostatic pressure had no apparent effect on uniform deformation in all cases, it still affected the deformation behavior after necking had occurred. Both a brass and copper failed by void sheeting, marked by an inclined planar fracture surface covered by
elongated shallow dimples, within shear zones at atmospheric pressure (0.1 MPa) and up to pressures between 300 MPa and 400 MPa.

Both materials fractured by necking down to a chisel edge as imposed hydrostatic pressure was increased further to 600 MPa. Figure 2.5 shows a schematic diagram of this process. Firstly, a broad neck is developed across the wide face of the sample (a) followed by localization into a single shear band (b). This band would work harden, and further deformation proceed by the formation of a second symmetrical shear band (c). This process repeated a large number of times until the sample necked down to a chisel edge and failed.

Figure 2.5 Schematic diagram of shear localization process in a sheet tensile sample (Weinrich and French, 1976).

According to their experimental results, the occurrence of shear localization is a function of the material itself and the sample geometry. This was exemplified by the fact that at low pressures, round samples of both α brass and copper exhibited a classic cup and cone feature involving void growth and coalescence, whereas sheet samples failed by
void sheeting within the shear band. In cylindrical geometry at sufficiently high pressure, brass failed by shear while copper failed by ductile rupture.

Chung et al. (1977) suggested that negative strain rate sensitivity can induce conditions for shear localization. Tensile tests were performed on cylindrical samples of 7075 Al-Zn-Mg alloy at 300 K and 77 K. They proposed that the occurrence of strain localization depends upon the attainment of the following condition:

\[
\left( \frac{\partial \sigma}{\partial \varepsilon} + \frac{\partial \sigma}{\partial \dot{\varepsilon}} \right) < 0 \quad \cdots (2.8)
\]

which requires either strain or strain rate softening processes. At 300 K the material had negative strain rate sensitivity and failure by shear localization occurred before the onset of necking. The material, in contrast, exhibited slight positive strain rate sensitivity at 77 K and necking occurred before shear localization.

### 2.4.2. Shear Localization in Pure Al and Al-Mg Alloys

The shear localization process often leads to the formation of an intense shear deformation region, so-called shear band. Shear band formation has been observed in both single crystal and polycrystal materials. Extensive dislocation scale microscopic examinations have been performed in order to correlate the occurrence of shear bands with microscopic features. Given the focus of the present work, attention will be given to investigations of shear band development in pure Al and Al-Mg alloys.

The formation of shear bands in Al and Al-3 wt.% Mg single crystals was examined by Morii et al. (1985). The Al-3 wt.% Mg single crystal with orientations
(211)\{11\} and (011)\{100\} and Al single crystal with orientation (211)\{11\} were rolled at room temperature. The shear bands can only be observed in Al-Mg crystal with orientation (211)\{11\}, shown in Figure 2.6. Transmission electron microscopy observations showed that a heterogeneous dislocation structure composed of well developed lamellae-like dislocation walls was found in (211)\{11\} Al-Mg crystal, whereas (211)\{11\} Al crystal formed equiaxed cell structure, instead. Therefore, they concluded that the occurrence of shear bands was significantly influenced by crystal orientation and closely correlated with the development of layered dislocation structure.

Brown (1972) reported shear band formation in 5657 Al-0.8 wt.% Mg. The material was rolled up to 50% strain and then its surface was polished. The material was then pulled further by 1% to 10% strain, and both shear band offsets and spacing between shear bands were measured. It was observed that 30% to 50% of the overall plastic deformation resulted from localized shear. In addition, shear band formation followed the predictions of continuum mechanics rather than being dependent upon crystallographic orientation of the grains.
Figure 2.6  Slip markings after room temperature rolling: (a) $(011)[\overline{1}00]$ Al-Mg rolled 70%, (b) $(211)[\overline{1}11]$ Al-Mg rolled 70% and (c) $(211)[\overline{1}11]$ Al rolled 80% (polished and additionally rolled about 10%). Shear bands can be seen only in crystal (b) (Morii et al., 1985).

A detailed study of shear band development in polycrystalline Al-4.8 wt.% Mg alloy was performed by Korbel et al. (1986). The material was deformed by rolling at room temperature and 77 K. Based on extensive transmission electron microscopy results, it was observed that the formation of a macroscopic shear band resulted from the propagation of micro-bands, which were crystallographic in character and formed on a variety of $\{111\}$ planes through the crystal. As these micro-bands developed into more macroscopic features, they broadened and were able to cross grain boundaries resulting in the spatial organization of local shear events. Moreover, the formation of macroscopic shear bands resulted from the propagation (growth) of micro-bands through the crystal,
not from a linkage of separate and independently formed micro-bands. The macroscopic shear bands were found to be a continuum rather than a crystallographic phenomenon, hence their overall orientation, as predicted by continuum plasticity model, lies approximately at \( \pm 35^0 \) towards the rolling plane. By comparing texture development in pure Al and Al-Mg alloy, it was concluded that shear banding occurs without texture softening.

Furthermore, Korbel and Martin (1988) examined the effect of strain path change toward the formation of shear bands in Al-killed steel containing 0.06 wt.% C. The materials were deformed in a wide range of strain from 0.1 to 0.9 of true rolling strain, and then tested in tension mode. Similar results were obtained, in which macroscopic shear bands originated from crystallographic micro-bands followed by their propagations across several grains. The sequential series of events leading to the formation of macroscopic shear bands are described schematically in Figure 2.7. The change of deformation path, from rolling to tension, considerably enhances the formation of micro-bands.

![Figure 2.7 Schematic visualization of the development of macroscopic shear band (Korbel and Martin, 1988).](image)
Bird et al. (1987) investigated the development of shear bands in both crystal scale and sample scale in 5182 Al-4.4 wt.% Mg under different stress state, i.e. plane strain conditions. The crystal scale and sample scale shear bands referred to micro-bands and macroscopic shear bands, respectively, were observed by Korbel et al. (1986). The crystal scale shear bands, which had been formed by transgranularly coordinated coarse slip, extended across a few grains once through-thickness necking occurred. As deformation continued, sample-scale shear bands initiated in regions of the sheet where crystal scale shear bands had already existed.

More recently, Fjeldly et al. (2001) studied strain localization in heat treated Al-Zn-Mg AA7030 alloy by performing uniaxial tensile tests and simple shear tests at room temperature. They found that the deformation structure in this alloy was heterogeneous from the sample scale to the dislocation scale. On the sample scale, there are macroscopic shear bands, which cause the final fracture. On the grain level, there are coarse slip bands, which grow and coalesce into crystallographic microbands. Finally, on the dislocation scale, there is a network of entangled dislocations cut through by narrow parallel bands of localized slip.

2.5. Digital Image Correlation (DIC) Technique

Digital image correlation (DIC) was originally introduced in the early 80’s by one research group of the University of South Carolina (Sutton et al., 1983). DIC is a non-contact optical method which uses image recognition to analyze and compare digital images acquired from the surface of the sample for measurement of displacement and
strain. A typical experimental setup of DIC is shown in Figure 2.8. Two incoherent light sources, which are utilized to uniformly illuminate the sample surface, are positioned on both sides of a CCD camera, which is placed normal to the sample surface. Two images are acquired; each is taken respectively before and after the deformation that results from mechanical loads. The two images are digitized and divided into several small sub-windows, so-called facets, and then stored on a computer for further processing.

With proper surface preparation, the interaction between the incident light and the sample surface results in an image with random, high contrast, and high frequency variations in light intensity; in other words, the surface appears “speckled”. Digital images acquired before and after deformation document the “speckle” distribution and can be represented by the gray-scale intensity distribution. The light intensity distribution at each point on the surface is unique and hence the distribution in light intensity about a particular point \((x,y)\) can be described by the grayscale matrix \(F(x,y)\) over a selected facet.
of the digital image. After deformation of the sample, each position of the surface \((x, y)\) is assumed to occupy a new location \((x^*, y^*)\).

The in-plane surface displacement can be determined by finding the position of the light intensity distribution \(F^*(x^*, y^*)\) that most closely resembles the original distribution \(F(x, y)\). A search is performed to find the location on the deformed image (after deformation) with the gray scale distribution that is most consistent with that on the original undeformed image. The location of \(F^*(x^*, y^*)\) can be obtained by finding the position with maximum correlation coefficient \((C)\) according to:

\[
C = \frac{\sum_{i,j=1}^{n} F(x_i, y_j) \cdot \sum_{i,j=1}^{n} F^*(x_i^*, y_j^*)}{\left[\sum_{i,j=1}^{n} F^2(x_i, y_j) \cdot \sum_{i,j=1}^{n} F^2(x_i^*, y_j^*)\right]^{\frac{1}{2}}} \quad \cdots (2.9)
\]

where \(F\) and \(F^*\) are the grey scale matrices of the facet at position \((x_i, y_j)\) in the undeformed image and \((x_i^*, y_j^*)\) in the deformed image, respectively. The values of \(C\) are confined by the inequality, \(0 < |C| < 1\), of which \(C = 1\) corresponds to perfect correlation.

A schematic description of the principle of DIC technique is shown in Figure 2.9.
The surface displacements are determined from the positions of the gray scale distribution of the deformed image with respect to that of the original image. If the out-of-plane displacement is small and can be neglected, the location in the deformed image \((x^*, y^*)\) is described by:

\[
x^* = x + u + \frac{\partial u}{\partial x} \Delta x + \frac{\partial u}{\partial y} \Delta y \quad \ldots \quad (2.10)
\]

\[
y^* = y + v + \frac{\partial v}{\partial x} \Delta x + \frac{\partial v}{\partial y} \Delta y \quad \ldots \quad (2.11)
\]

where \(u\) and \(v\) are the displacements of the facet centre in the \(x\) and \(y\) directions, respectively. The terms \(\Delta x\) and \(\Delta y\) are the distances from the facet centre to point \((x,y)\).

In traditional DIC, the quantities \(u, v, \frac{\partial u}{\partial x}, \frac{\partial u}{\partial y}, \frac{\partial v}{\partial x}, \frac{\partial v}{\partial y}\) are evaluated iteratively and determined from the position with maximum correlation coefficient. Later
on, Bruck et al. (1989) developed a new digital image correlation algorithm which was based on the Newton-Raphson method of partial differential correction. This algorithm allowed for a reduction in computation time for calculations of displacements and strains.

2.6. Assessment of the Literature

It has been generally accepted that PLC effect is caused by dynamic interactions between diffusing solute atoms and mobile dislocations throughout the deformation process. The PLC effect has been experimentally observed in both interstitial and substitutional alloys within a certain range of temperature and strain rate. Classifications have been made based on serration characteristics of the stress-strain curve, which are closely related to the dynamic behavior of the corresponding deformation bands. Both numerical modeling and finite element simulations have been extensively developed in order to describe the occurrence of PLC effect.

Shear band formation and development have been observed experimentally in both pure Al and Al-Mg alloys. Many TEM studies have shown that the presence of macroscopic shear bands are correlated with the formation of crystallographic grain-scale shear bands or micro shear bands during an early stage of the deformation process. As plastic strain increases, the grain-scale shear bands propagate across several grains to form a non-crystallographic macroscopic shear band.

On one hand, Spencer et al. (2002) has suggested that 5754 Al-Mg alloy fails by shear banding which limits the use of this alloy for automotive applications. On the other hand, Kang et al. (2005) has observed the impact of PLC banding, specifically type B
bands, towards a reduction in ductility of the same alloy. These facts have motivated us to study the relationship between these two different phenomena occurring in the 5754 Al-Mg alloy in order to optimize the formability of this alloy. In this work, the impact of PLC banding on shear banding is studied by performing uniaxial tensile tests coupled with non-contact strain measurement, so-called DIC technique. In addition to that, study of PLC banding as a function of prior deformation under positive strain rate sensitivity is explored.
CHAPTER 3

EXPERIMENTAL PROCEDURE

3.1. Materials

The material investigated in this research is the AA 5754 Al-Mg alloy, which was received in the form of 1 mm thick cold-rolled sheet with typical composition shown in Table 3.1. The material was annealed for 2 hours at 350°C and air cooled resulted in an “O” temper condition. After annealing process, the material was then held at room temperature for at least 2 weeks prior to tensile testing. In addition to the principal alloying element Mg, the alloy also contains Mn, Si, and Fe. The alloy is classified as non-heat treatable.

Table 3.1 Chemical composition of 5754 Al-Mg alloy in weight %.

<table>
<thead>
<tr>
<th></th>
<th>Mg</th>
<th>Mn</th>
<th>Si</th>
<th>Fe</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>AA 5754</td>
<td>3.50</td>
<td>0.21</td>
<td>0.11</td>
<td>0.21</td>
<td>balance</td>
</tr>
</tbody>
</table>

3.2. Uniaxial Tensile Test

Tensile samples were machined in the rolling direction according to Figure 3.1. All dimensions shown are in millimeter. All tests were performed at constant crosshead speed, which corresponds to an initial bulk strain rate of $6 \times 10^{-4}$ s$^{-1}$. 
3.2.1. Tensile Test at Room Temperature

Tensile tests at room temperature (298 K) were carried out using a screw-driven tensile machine, Instron 5566. Extension of the gauge section was measured using an extensometer with 12.5 mm gauge length. Non-contact strain measurement with digital image correlation (DIC) technique, which enabled observation of PLC band behavior, was also performed. Details of this technique will be discussed in the next section.

3.2.2. Tensile Test at 223 K

Tensile tests at 223 K were performed by immersing the samples in a styrofoam chamber containing a mixture of ethanol and dry ice. Due to non-compatibility of the chamber with Instron machine, tests at 223 K were performed by using a servo-hydraulic universal MTS 880 machine. Some of room temperature tests were performed in MTS frame, and then the results were compared to those obtained by Instron machine. There is
no marked difference observed, which suggests that machine-dependent effect can be neglected. The experimental setup is shown in Figure 3.2. Strain measurements were carried out using a clip-on extensometer with 12.7 mm gauge length.

Figure 3.2  Low temperature tensile test setup of a servo-hydraulic universal MTS 880 machine utilized in the experimental study.
3.2.3. Pre-straining Experiment

In order to assess PLC banding as a function of prior deformation under positive strain rate sensitivity condition, tensile samples were deformed up to several different levels of true strain at 223 K and then unloaded. A speckle pattern was applied on the surface of the sample prior to reloading at room temperature to enable strain measurement by the DIC technique. There was always approximately half an hour gap between low temperature test and subsequent room temperature test, which was required for unloading, sample preparation, and reloading.

3.2.4. Pre-straining Experiment with Aging Treatment

Tensile samples, which had previously been deformed up to a true strain of 0.15 at 223 K, were used to study the effect of aging treatment on PLC banding. Aging treatment of the deformed samples was then performed at room temperature with two different aging periods of 50 and 120 hours.

3.3. Non-contact Strain Measurement by DIC Technique

DIC technique was utilized in order to observe the behavior of the plastic deformation band, so-called PLC band, during tensile tests. A typical experimental setup for non-contact strain measurement with DIC technique is shown in the previous chapter (see Figure 2.8). Basically, strain measurement using DIC technique involves three main steps, namely:
1. Sample Preparation

All sample surfaces were mechanically lapped with 600 grade SiC paper to a smooth finish, free of any machining marks. One flat surface of the tensile samples was decorated with fairly uniform random white and black speckles by a fine mist of water-based paint spray. Two primary features of the speckle field, which influence the robustness and precision of strain measurement by DIC technique, are the gray scale variation and the speckle size. The gray scale range between speckles and background should enable the digitized gray scale distribution of a facet to have its own unique distribution that can be recognized by the cross correlation function after deformation. The speckle size must be small enough in order to achieve a desired spatial resolution of measurement.

2. Data Acquisition

A CCD camera with resolution of $1280 \times 1024$ pixels was utilized to acquire digital images at a sampling rate of either 0.5 frame per second or 1 frame per second during tensile test. The distance between camera and sample surface is set approximately 40 cm in order to capture the whole gauge length of the sample. Care was taken in order to avoid optical errors which are related to focusing errors due to the sample deformation and sample illumination. The digital images acquired by CCD camera have a spatial resolution of about 0.025 mm per pixel.
3. Data Analysis

A commercially available digital image processing, ARAMIS, was used to calculate the displacement and ultimately the strain between images. The facet size used in calculation is $9 \times 9$ pixels which correspond to $0.225 \times 0.225$ mm of actual dimensions on sample surface.

3.3.1. Calibration Procedure

A calibration procedure was performed in order to assess the error level of the DIC measurement technique. One end of the sample was clamped to the moving end of the Instron machine, while the other end was left unclamped. After the first image was taken, the moving-end was shifted with a prescribed amount of displacement, and then the second image was acquired. Afterwards, the measured and calculated displacements were compared and analyzed. It was found that the error of this DIC measurement technique is less than 0.25%.

3.3.2. Band Strain Measurement

The amount of strain carried by a PLC band, so-called band strain, can be simply measured by further processing of DIC results. Figure 3.3 (a) shows an example of a strain map obtained from a DIC measurement. Note that the “[log]” scale bar represents the true strain value. Figure 3.3 (b) displays a profile of the strain levels along the line section drawn in Figure 3.3 (a). For clarity purposes, however, not all data points are plotted in Figure 3.3 (b). The band strain is simply obtained by subtracting the average
strain value from the maximum strain value within the band, in which the average strain is acquired by averaging all data points of true strain along the line section.

Figure 3.3  (a) A strain map obtained from DIC measurement is utilized for determination of PLC band strain, (b) The profile of strain levels for the line section drawn in (a). For clarity purposes, note that not all data points are plotted in the strain profile. The \( ['log'] \) scale bar represents the true strain value. The y-axis error bars are based on the maximum potential strain measuring error of 0.25\%.
3.4. Etching for Shear Bands

Fractured samples were sectioned down their centre line with a low speed carbide saw, and then mounted in Struers Epofix Resin. The samples were successively ground using finer grades of silicon carbide paper, followed by 3 \( \mu m \) and 1 \( \mu m \) diamond slurry. The final polishing step was carried out using 0.05 \( \mu m \) alumina paste. In order to reveal bands of intense shear, this procedure was followed by a dislocation decoration technique (Lloyd et al., 1982). The mounted samples were aged at 90\(^\circ\)C for 14 days. During the aging period, magnesium atoms will diffuse to high dislocation density regions, such as shear bands and grain boundaries, and form precipitates. The samples were then etched for 70 to 100 seconds in a solution of 10\% H\(_3\)PO\(_4\) held at 60\(^\circ\)C. This treatment preferentially etches the precipitates so that shear bands and grain boundaries will be outlined by fine etch pits.

3.5. Fracture Surfaces Observation

The fracture surfaces of broken samples were examined in a Philips 515 scanning electron microscope operating at an acceleration voltage of 20 keV.
CHAPTER 4

EXPERIMENTAL RESULTS

This chapter will present all experimental results obtained throughout this study. The various results in this chapter are classified as follows: (i) results of uniaxial tensile tests at both 298 K (room temperature) and 223 K, (ii) digital image correlation (DIC) results for all tensile tests performed at room temperature, (iii) band strain measurements based on DIC results, (iv) shear localization observation, and (v) fracture surfaces observations.

4.1. Uniaxial Tensile Tests

In order to characterize the Portevin-Le Chatelier (PLC) effect in the 5754 Al-Mg alloy, uniaxial tensile tests were conducted at both room temperature (298 K) and 223 K. The impact of the PLC effect on shear localization can be investigated by making use of pre-deformed tensile samples at 223 K, at which temperature the PLC effect is significantly suppressed, prior to room temperature tests. All tensile data are presented in terms of true stress (MPa) versus true strain curves. All tests were performed at least two times to ensure the repeatability of the experimental results obtained in this study.
4.1.1. Uniaxial Tensile Tests at 298 and 223 K

The uniaxial true stress-true strain curves for a sample deformed at room temperature at two different strain rates, i.e. \(6 \times 10^{-4} \text{ s}^{-1}\) and \(3 \times 10^{-3} \text{ s}^{-1}\), are shown in Figure 4.1. The 5754 Al-Mg alloy exhibits a distinct Lüders strain which indicates a low initial density of mobile dislocations. The presence of PLC effect is marked by a discontinuous yielding phenomenon during tensile test, resulting in a serrated stress-strain curve. At room temperature, the Al-Mg 5754 alloy exhibits the negative strain rate sensitivity behavior, where flow stress decreases with increasing strain rate. Due to the negative strain rate sensitivity, the flow curve at strain rate of \(3 \times 10^{-3} \text{ s}^{-1}\) is lower than the one obtained at \(6 \times 10^{-4} \text{ s}^{-1}\), as shown in Figure 4.1.

The amplitude of serrations was found to increase gradually with strain and then finally saturate at large strain, approximately \(\pm 7 \text{ MPa}\) about the mean flow stress at \(6 \times 10^{-4} \text{ s}^{-1}\) and \(\pm 4 \text{ MPa}\) at \(3 \times 10^{-3} \text{ s}^{-1}\). In addition, the amplitude of serrations decreases with decreasing strain rate, as shown in Figure 4.1. The serrations appearing in both stress-strain curves suggest that the type B serration is particularly preferred in strain rate range between \(6 \times 10^{-4} \text{ s}^{-1}\) and \(3 \times 10^{-3} \text{ s}^{-1}\) at room temperature.
Figure 4.1  True stress-true strain curves for sample tested at 298 K with two different strain rates. Both curves show characteristic of type B serration.

The PLC effect, which is commonly associated with the dynamic strain aging phenomenon, however, is significantly suppressed at 223 K. Consequently, a smooth stress-strain curve is obtained, as shown in Figure 4.2. The flow curve indicates that Lüdering still occurs at this temperature with a small increase in Lüders strain in comparison to that observed in the room temperature test at the same strain rate. Based on
the results of the strain rate change test, it was observed that the 5754 Al-Mg alloy exhibits positive strain rate sensitivity behavior at 223 K.

Figure 4.2 True stress-true strain curve for sample tested at 223 K with strain rate of $6 \times 10^{-4}$ s$^{-1}$. PLC effect is significantly suppressed, as manifested in smooth flow curve.

Tensile test data derived from both curves are summarized in Table 4.1. The strain hardening coefficient, $n$, is calculated according to the power–law hardening formula, $\sigma = K \varepsilon^n$. An error estimate for the $n$ value was obtained from standard deviation of a linear fit of log $\sigma$-log $\varepsilon$ curve using Microcal Origin software. Note that $n$ is essentially the same for all three testing conditions. The sample tested at 223 K shows a
significantly higher uniform elongation in comparison with those tested at room temperature.

Table 4.1  Tensile data for 5754 Al-Mg alloy tested at both 298 K and 223 K, in the rolling direction.

<table>
<thead>
<tr>
<th>T (K)</th>
<th>Strain Rate (s⁻¹)</th>
<th>Uniform Strain</th>
<th>True 0.2% Offset Yield Strength (MPa)</th>
<th>True Ultimate Tensile Strength (MPa)</th>
<th>Strain Hardening Exponent, n</th>
</tr>
</thead>
<tbody>
<tr>
<td>298</td>
<td>6 × 10⁻⁴</td>
<td>0.205</td>
<td>147</td>
<td>326</td>
<td>0.30 ± 0.02</td>
</tr>
<tr>
<td>298</td>
<td>3 × 10⁻³</td>
<td>0.215</td>
<td>138</td>
<td>310</td>
<td>0.30 ± 0.02</td>
</tr>
<tr>
<td>223</td>
<td>6 × 10⁻⁴</td>
<td>0.24</td>
<td>153</td>
<td>340</td>
<td>0.31 ± 0.02</td>
</tr>
</tbody>
</table>

4.1.2. Pre-straining Experiment

The results of these experiments will be presented in two sub-sections based on the given amount of pre-strain at 223 K. A series of samples with small amount of varying pre-strain, namely 0.02, 0.1, and 0.15 of true strain, were deformed to study PLC banding as a function of prior deformation under positive strain rate sensitivity conditions. As the amount of pre-strain at 223 K is increased up to a point which is larger than the necking strain at room temperature, the impact of PLC banding towards the final shear localization process can be carefully assessed.

4.1.2.1. Low Range of Pre-strain

Typical true stress-true strain curves obtained for samples with low range of pre-strain are described in Figure 4.3 to 4.5. During reloading at room temperature, plastic deformation takes place as soon as the current flow stress is slightly higher than
maximum flow stress which was reached on previous deformation. Simultaneously, the PLC effect reappears as manifested by the presence of load instability in the stress-strain curve. An increase of magnitude of serrations with increasing strain is observed in all samples. It is also noticed that, at the same level of strain, the magnitude of serrations is comparable with that found in the sample with no pre-strain. The work hardening behavior does not seem to be affected by prior deformation at 223 K, as the flow stress curve continues to show the similar trend.

Figure 4.3 Sample was deformed up to true strain of 0.02 at 223 K, unloaded, and then reloaded at 298 K. Serrations reappeared during reloading at 298 K.
Figure 4.4  Sample was deformed up to true strain of 0.1 at 223 K, unloaded, and then reloaded at 298 K. Serrations reappeared during reloading at 298 K.
Sample was deformed up to true strain of 0.15 at 223 K, unloaded, and then reloaded at 298 K. Serrations reappeared during reloading at 298 K.

**4.1.2.2. High Range of Pre-strain**

A true stress-true strain curve for a sample deformed at 223 K up to true strain of 0.235 followed by further straining at room temperature until fracture is shown in Figure 4.6. Note that this is above the uniform strain that was achieved in the room temperature tests (see Table 4.1). A plateau part of the flow curve is observed during reloading at room temperature, followed by small work hardening prior to fracture.
Sample was deformed up to true strain of 0.235, just before necking occurs, at 223 K and further deformed until fracture at room temperature.

Figure 4.7 shows a true stress-true strain curve for a sample deformed at 223 K up to necking point, i.e. true strain of 0.24, and then further deformed at room temperature until fracture. During reloading at room temperature, plastic deformation occurs at slightly lower flow stress in comparison to maximum flow stress reached on previous deformation.
Sample was deformed up to necking point at 223 K, i.e. true strain of 0.24, and further deformed until fracture at room temperature.

4.1.3. Pre-Straining Experiments with Aging Treatment

In this type of experiment, all samples were deformed up to true strain of 0.15 at 223 K and then aged at room temperature with two different aging time, namely 50 hours and 120 hours. Afterwards, the samples were pulled until fracture at room temperature with the same crosshead velocity. The corresponding true stress-true strain curves are shown in Figure 4.8. During reloading at room temperature, both samples start to deform
at lower flow stress level in comparison with previous flow stress. The longer the aging time, the lower the stress needed for plastic deformation to take place. This decrease of flow stress persists only for approximately four percents of strain and then both curves continue to show similar trend with the one for low temperature curve. Nevertheless, the increase of magnitude of serrations with increasing strain is still maintained.

Figure 4.8 True stress-true strain curves for sample pre-strained up to true strain of 0.15 at 223 K, aged for 50 hours and 120 hours at 298 K, and then retested at 298 K.
4.2. Digital Image Correlation (DIC) Results

Observations of PLC band behavior were carried out for all tensile tests performed at room temperature. All digital images were captured with a sampling rate of 0.5 frames per second, unless specified. Figure 4.9 shows a series of strain maps obtained by the DIC technique for a sample tested at room temperature with a strain rate of $6 \times 10^{-4} \, \text{s}^{-1}$. Characteristics of type B PLC bands, where deformation bands are nucleated at random locations followed by discontinuous propagation, can be clearly seen in the two series of strain maps. Multiple bands appear simultaneously at different locations along the gauge length of the sample. These bands travel only a short distance, some of them do not show any movement at all. These bands appear at a fixed angle ranging from $55^0$ to $60^0$ with respect to the tensile axis. The alteration of inclination angle during the test was also observed. Type B PLC banding is also observed in samples tested at room temperature with a strain rate of $3 \times 10^{-3} \, \text{s}^{-1}$.

DIC results obtained for all pre-strained samples will be presented in three subsections based on the extent of pre-strain at 223 K, i.e. a low range and a high range of pre-strain, and aging treatment.
Figure 4.9  Strain map for sample tested at room temperature with strain rate of $6 \times 10^{-4}$ s$^{-1}$. Type B PLC bands are observed which marked by random nucleation of deformation bands.

Figure 4.10  Strain map for sample pre-strained up to true strain of 0.02 at 223 K with strain rate of $6 \times 10^{-4}$ s$^{-1}$ showing characteristics of type B PLC bands. Note that the strain indicated in this figure does not include the amount of pre-strain at 223 K.
4.2.1. Low Range of Pre-strain

The following figures summarize the results of DIC measurements for samples pre-strained up to true strains of 0.02, 0.1, and 0.15, respectively (Figure 4.10, 4.11, and 4.12). Type B band behavior, typically found in samples without pre-strain, was also observed in a sample pre-strained up to true strain of 0.02.

Both series of strain maps for samples pre-strained up to true strain of 0.1 and 0.15 indicate a change in the nature of PLC banding in comparison to that observed in the sample without pre-strain. Figure 4.11 shows the behavior of the first deformation band formed during reloading at room temperature. The propagation of the band in pre-strained samples is more distinct in comparison to that in sample without pre-strain. The band continues to propagate until new bands nucleate at other locations along the gauge length, as shown in Figure 4.12. This sequence repeatedly occurs until the static shear band is formed. The inclination angle of the band in respect to the tensile axis, however, is not affected by prior deformation at 223 K.
Figure 4.11 Strain map for sample pre-strained up to true strain of 0.1 at 223 K with strain rate of $6 \times 10^{-4}$ s$^{-1}$ showing a distinct propagative band. Note that the strain indicated in this figure does not include the amount of pre-strain at 223 K.

Figure 4.12 Strain map for sample pre-strained up to true strain of 0.15 at 223 K with strain rate of $6 \times 10^{-4}$ s$^{-1}$ showing a distinct propagative band. The band continues to propagate until new band nucleates at other locations. Note that the strain indicated in this figure does not include the amount of pre-strain at 223 K.
4.2.2. High Range of Pre-strain

The behavior of deformation bands in a sample pre-strained to a true strain of 0.235, i.e. just before necking point, at 223 K is described in Figure 4.13. During reloading at room temperature, the first deformation band nucleated at one end of the sample, and then continuously propagated to the other end. This feature is followed by the formation of a final shear localization band on another plane, which crosses the sample in roughly the opposite direction.

DIC results obtained for a sample pre-strained up to necking point at 223 K (Figure 4.14) shows that no PLC band appeared during reloading at room temperature. Plastic yielding leads directly to the formation of the first shear band, as shown in Figure 4.14 (a). Further deformation is accommodated by formation of the conjugate band. Finally, an intersection of those two shear bands triggers an initiation of fracture of the sample.
Figure 4.13 Strain map for sample pre-strained up to true strain of 0.235 at 223 K with strain rate of $6 \times 10^{-4} \text{ s}^{-1}$. The formation of shear band takes place on another plane, which crosses the sample in roughly the opposite direction. Note that the strain indicated in this figure does not include the amount of pre-strain at 223 K.
Figure 4.14 Strain map for sample pre-strained up to necking point at 223 K with strain rate of $6 \times 10^{-4} \text{s}^{-1}$. There is no PLC band appeared and plastic yielding leads directly to the formation of the first shear band. Note that the strain indicated in this figure does not include the amount of pre-strain at 223 K.

4.2.3. Pre-strained Samples with Aging Treatment

The strain maps for samples pre-strained up to true strain of 0.15 at 223 K and then aged for 50 and 120 hours, prior to room temperature testings are shown in Figure 4.15 and 4.16. A distinct propagative deformation band was observed in a sample that was aged for 50 hours (Figure 4.15), whereas a sample aged for 120 hours showed random nucleation of deformation bands across the gauge length (Figure 4.16).
Figure 4.15 Strain map for sample pre-strained up to true strain of 0.15 at 223 K with strain rate of $6 \times 10^{-4}$ s$^{-1}$, and then aged for 50 hours at 298 K prior to room temperature testing. Note that the strain indicated in this figure does not include the amount of pre-strain at 223 K.

Figure 4.16 Strain map for sample pre-strained up to true strain of 0.15 at 223 K with strain rate of $6 \times 10^{-4}$ s$^{-1}$, and then aged for 120 hours at 298 K prior to room temperature testing. Note that the strain indicated in this figure does not include the amount of pre-strain at 223 K.
4.3. Band Strain as a Function of Global Strain

Further processing of strain maps obtained by the DIC technique was performed in order to observe the development of band strain as a function of global strain. Band strain is defined as the extent of incremental strain carried by the deformation band, while global strain represents the strain measured by extensometer. A comparison of the development of band strain in two samples tested at room temperature with different strain rates is summarized in Figure 4.17. Both results show that band strain increases linearly with global strain.

![Graph showing band strain as a function of global strain](image)

**Figure 4.17** A plot of band strain as a function of global strain in samples tested at room temperature with different strain rates. Band strain increases linearly with global strain.
The results of band strain measurement in all pre-strained samples are summarized in Figure 4.18. Linear relationship between band strain and global strain is still observed.

![Figure 4.18: A linear relationship between band strain and global strain in pre-strained samples. All tensile tests were performed with strain rate of $6 \times 10^{-4} \text{s}^{-1}$. Note that the global strain plotted in the curve includes the amount of pre-strain at 223 K.](image)

The effect of aging treatment on the development of band strain is described in Figure 4.19. Although band strain starts off much lower value than for samples with the
same amount of pre-strain and without aging, the linear relationship between band strain and global strain is still observed in aged samples. Band strain of aged samples, however, increases faster with increasing global strain in comparison to those measured in samples tested at room temperature.

![Figure 4.19](image)

Figure 4.19 A plot of band strain as a function of global strain in sample tested at room temperature and aged samples. All tensile tests were performed with strain rate of $6 \times 10^{-4} \text{s}^{-1}$. Note that band strain in aged samples starts off much lower value.
4.4. Shear Localization Observation

Investigations of necked region were made for samples fractured at both room temperature and 223 K. All samples have been aged and etched in order to reveal the presence of intense shear bands. At both temperatures, the 5754 Al-Mg alloy shows the presence of extensive shear band formation within the necked region, as shown in Figure 4.20 and Figure 4.21. However, sample tested at 223 K shows higher total ductility, as manifested in larger thickness reduction.
Figure 4.20 Fracture of sample tested at room temperature, aged to reveal intense shear bands.
4.5. Fracture Surfaces Observation

The fracture details of tensile samples tested at both room temperature and 223 K were observed using scanning electron microscopy. The shallow and elongated dimples, as shown in Figures 4.22 and 4.23, suggest that the 5754 Al-Mg alloy fails in shear by void sheeting at both temperatures. This finding is in accordance with the shearing behavior study presented in the previous section, in which intense shear bands were revealed by etching. Moreover, fracture surfaces were found to be very fibrous, indicating a large amount of damage in the form of void nucleation and growth prior to fracture.
Figure 4.22  SEM fractographs of tensile sample pulled until fracture at room temperature, which show shallow and elongated dimples.
Figure 4.23 SEM fractographs of tensile sample pulled until fracture at 223 K, which show shallow and elongated dimples.
CHAPTER 5

DISCUSSION

This chapter will discuss the experimental results presented in Chapter 4 in order to fulfill the following objectives: (i) to compare work hardening behavior at different strain rates and temperatures, (ii) to describe the PLC banding as a function of prior deformation under positive strain rate sensitivity, (iii) to implement Kang’s model (2005) in predicting necking strain based on band strain measurement, (iv) to analyze the recovery process induced by aging treatment and its effect on band strain development, (v) to evaluate the impacts of PLC banding on shear banding, and finally (vi) to discuss shear banding as a potential failure mode in 5754 Al-Mg alloy.

5.1. Work Hardening Behavior at 223 K and 298 K

The work hardening rates of the 5754 Al-Mg alloy at room temperature can be found from discrete measuring points constituting the upper profile of serrations of the true stress-true strain curve. Based on the Considère criterion, diffuse necking in sheet tensile samples should take place when the work hardening rate of a material equals its corresponding true stress, i.e. \( \frac{d\sigma}{d\varepsilon} = \sigma \). The 5754 Al-Mg alloy at room temperature, however, does not follow this criterion. Instead, as shown in Figure 5.1, diffuse necking commences even when the work hardening rate of the alloy is considerably higher in
comparison to the corresponding true stress. An early onset of necking during uniaxial tension was also observed in the 5182 Al-Mg alloy by Tong (1997). He suggested that the initiation and growth of necks may be attributed to interplay of strain hardening, strain inhomogeneity, and surface topography during deformation process.

![Graph showing work hardening behavior of 5754 Al-Mg alloy at room temperature.](image)

**Figure 5.1** Work hardening behavior of 5754 Al-Mg alloy at room temperature.

By incorporating the Considère criterion into Hollomon equation, $\sigma = K\varepsilon^n$, a simple relationship for the strain at which diffuse necking should occur can be obtained (Equation 5.1). The corresponding strain is the maximum uniform strain, $\varepsilon_u$, and $n$ represents the strain hardening coefficient.
\[ \varepsilon_u = n \quad \ldots (5.1) \]

Due to the occurrence of the PLC effect, however the necking strain at room temperature is found to be considerably lower than strain hardening exponent of 0.30.

Figure 5.2 shows that there is very little discernable variation of the work hardening behavior at different test temperatures, for a given strain rate of \( 6 \times 10^{-4} \text{ s}^{-1} \), up to the point of diffuse necking. However, at 223 K, the PLC effect is significantly suppressed and the work hardening rate at the onset of diffuse necking lies very close to the flow stress. In other words, diffuse necking does not occur until the Considère criterion is met. Table 5.1 summarizes the work hardening rate and its corresponding true stress when the diffuse necking commences. This comparison suggests that PLC effect, as a result of dynamic strain aging phenomenon, causes a decrease in ductility of 5754 Al-Mg alloy by triggering an early onset of diffuse necking during the uniaxial tensile test. In addition, when the PLC effect exists, it seems that the fracture process is not related to exhaustion of work hardening capacity, as is usually observed during uniform elongation.

Table 5.1 Work hardening rate and its corresponding true stress at a point of diffuse necking.

<table>
<thead>
<tr>
<th>Temperature (K)</th>
<th>Strain Rate (s(^{-1}))</th>
<th>Necking Strain</th>
<th>Strain Hardening Exponent, n</th>
<th>At Necking Point</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Work Hardening Rate (MPa)</td>
</tr>
<tr>
<td>298</td>
<td>(6 \times 10^{-4})</td>
<td>0.205</td>
<td>0.30 ( \pm ) 0.02</td>
<td>471</td>
</tr>
<tr>
<td>298</td>
<td>(3 \times 10^{-3})</td>
<td>0.215</td>
<td>0.30 ( \pm ) 0.02</td>
<td>433</td>
</tr>
<tr>
<td>223</td>
<td>(6 \times 10^{-4})</td>
<td>0.24</td>
<td>0.31 ( \pm ) 0.02</td>
<td>361</td>
</tr>
</tbody>
</table>
5.2. Spatial Coupling of Type B PLC Bands

During jerky flow of dislocations, it is generally accepted that the motion of dislocations, by which plastic deformation is carried out, can be seen as a sequence of free-flight and waiting events. At the dislocation scale, the appearance of a plastic strain burst in a form of a deformation band, known as a PLC band, is associated with the
collective motion of mobile dislocations, which are temporarily held up at localized obstacles in the glide plane. According to Sleeswyk (1958) and Mulford and Kocks (1979), the localized obstacles are likely to be forest dislocations. This dislocation scale theory, however, cannot describe the spatial characteristics of different types of PLC band as the applied strain rate or test temperature is varied.

In order to best describe the transition from randomly nucleated static bands (type C) to hopping propagating bands (type B) and finally to continuously propagating bands (type A), the spatial coupling between various volume elements of the sample has to be introduced. One possibility is to look at the coupling mechanism due to elastic stresses associated with local plastic strain incompatibilities (Lebyodkin et al., 1996). The DIC measurements show that plastic strain within the PLC band is always higher in comparison to the global strain. Therefore, based on the analysis given by Eshelby (1961), the internal stress will be developed in a material volume that experiences more plastic strain than the neighbouring material volume. The presence of internal stresses associated with non-uniformities of plastic strain provides a spatial coupling tending to equilibrate the plastic strain throughout the sample.

When the PLC effect occurs, relaxation of the internal stresses is observed (Chihab et al., 1987). Thus, both the magnitude and spatial extent of these internal stresses are expected to decrease with time once a localized deformation band is formed. At very low strain rates, the reloading time between two successive stress drops is very large. There is enough time for internal stresses to relax completely, hence no spatial correlation between bands is observed. This corresponds to the occurrence of type C
bands. At higher strain rates, where type B bands are observed, the reloading time decreases and becomes commensurate with the relaxation time. Therefore, internal stresses are not totally relaxed and favor the nucleation of a new band nearby the previous one. This describes the hopping characteristic associated with type B bands.

5.3. PLC Banding as a Function of Prior Deformation under Positive Strain Rate Sensitivity Condition

As shown in strain maps obtained by the DIC technique (presented in Figure 4.11 and Figure 4.12), pre-deformation of 5754 Al-Mg alloy at 223 K prior to room temperature testing causes a change in the nature of PLC banding. Random nucleation of deformation bands is not observed in the pre-strained samples; instead propagation of a single band becomes more prominent. Although the band behavior in pre-strained sample is different from that observed in sample without pre-strain, there is no apparent difference in terms of serration characteristics of the stress-strain curves. This suggests that pre-deformation at 223 K causes a change at the microstructural level, which is manifested in how deformation bands behave.

In order to examine a change in the nature of PLC banding in a more detailed manner, a DIC measurement was performed with a higher sampling rate, i.e. five frames per second. The corresponding results of this DIC measurement, presented in Figure 5.3, suggests the existence of two subsequent periods associated with band movement, namely an incubation period and a propagation period. The incubation period is indicated by an absence of deformation activity both outside and inside the band, as shown in
Figure 5.3 (a). The inactivity during this incubation period is clearly confirmed by the axial strain distribution profile in Figure 5.4, in which the strain profile remained constant, indicating that there is no major motion of band strain distribution along the tensile direction.

Tong et al. (2005) utilized high-speed digital photography and image correlation technique to observe the behavior of individual type-B PLC bands in 5052 Al-Mg sheet. Due to high temporal resolution of the digital photography system used in their experimental setup, which is approximately 0.2 ms, they were able to characterize the nucleation and growth of individual PLC bands during a tensile test. They found that PLC banding was not first fully developed to its peak strain level, but instead there was a gradual increment of plastic strain during band growth. Moreover, their strain measurement results did not show any significant expansion of the band along the width direction during its growth. Our results are in favor with their observation, however, we were not able to capture the plastic strain development during band growth due to the limit of temporal resolution in our strain measurement system.
Figure 5.3 Strain map showing (a) incubation period and (b) propagation period of PLC band in pre-strained sample. The sample was pre-strained up to true strain of 0.15 at 223 K prior to room temperature testing. Note that the strain shown in scale bar does not include the amount of pre-strain.
Figure 5.4  True axial strain distribution along the centerline of the pre-strained sample during incubation period. Note that the strain shown in y-axis does not include the pre-strain at 223 K.

A series of strain maps collected during the propagation period is presented in Figure 5.3 (b). The band propagates in the way that the trailing edge of the band stops deforming, which is possibly due to solute locking, and then dislocations activity starts to appear at the band front. The band continues to propagate until reaching one end of the sample. As soon as one band stops propagating, another band will nucleate at other locations. Moreover, the strain distribution profiles recorded during propagation period,
as shown in Figure 5.5, indicate a relatively constant extent of strain within the propagating band. Therefore, it can be concluded that dislocations activity at the band front involves a constant density of dislocations.

![True axial strain distribution along the centerline of the pre-strained sample showing constant band strain during propagation period. The strain shown in y-axis does not include the amount of pre-strain at 223 K](image)

Figure 5.5

As a result of the occurrence of the PLC effect right after Lüdering, the distribution of solute atoms and dislocations at room temperature becomes inhomogeneous. Such a condition triggers a random nucleation of deformation bands
with short-distance propagation. In contrast, during the deformation process at 223 K, during which the PLC effect is suppressed, solute atoms in the material do not segregate into temporarily arrested mobile dislocations. This leads to a more homogeneous distribution of solute atoms and dislocations than that in a sample deformed at room temperature. Consequently, once a deformation band is formed, it will be more likely to propagate rather than to nucleate another new band at other location. Furthermore, the existence of a stress gradient between deformed and undeformed regions at the band front also provides additional driving force for band propagation.

The above-mentioned finding implies that if a given pre-deformation in the sample at 223 K is not sufficient to develop a homogeneous distribution of solute atoms and dislocations, then there is no difference in PLC banding characteristics with the non pre-strained sample. This statement is confirmed by the results obtained from the pre-straining experiment, at which the sample was pre-deformed up to true strain of 0.02, just after Lüders strain, at 223 K. The typical characteristic of type B PLC band, commonly found in samples without pre-strain, was observed in this sample.

Moreover, the experimental results of aged pre-strained samples also support this explanation. The characteristic of type B PLC bands was observed in pre-strained sample aged for 120 hours (presented in Figure 4.16), whereas a distinct propagative PLC band still appeared in pre-strained sample aged for 50 hours (presented in Figure 4.15). On one hand, even though the extent of pre-strain is sufficiently large to develop a homogeneous distribution of solute atoms and dislocations, the 120 hours aging treatment provides sufficient time and driving force for solute atoms to diffuse to dislocations and build
solute atmosphere around them. On the other hand, although diffusion process takes place during 50 hours aging treatment, the solute atoms experience lack of time to completely build solute atmosphere around dislocations. This explains the difference in a nature of PLC banding due to different aging time.

During reloading at room temperature, as soon as the stress needed by dislocations to break away from their solute atmosphere is reached, a plastic strain burst in a form of localized deformation band commences. As the flow stress continues to rise, dislocations at other locations will be released to carry out further deformation. This condition brings back the random nucleation characteristic of deformation banding associated with type B PLC bands, as observed in pre-strained sample aged for 120 hours. Occurrence of dynamic strain aging, which gives rise to inhomogeneous distribution of solute atoms and dislocations, at room temperature also supports reappearance of type B bands in aged pre-strained samples.

5.4. Determination of Necking Strain based on Band Strain Measurement

Kang et al. (2005) has developed a model to quantify the effect of type B PLC bands towards an early onset of necking in uniaxial tension. In this model, PLC bands are considered to represent geometric imperfections in the tensile sample. This model makes use of the analysis given by Marciniak et al. (2002) in order to calculate a difference between the strain at maximum load in an imperfect sample, which represents the sample containing PLC bands, and the one in a perfect sample. For clarity purposes, a full
derivation of the model is presented in the Appendix. In brief, the necking strain in a sample containing PLC bands, \( \varepsilon_u \), can be calculated as follows:

\[
(n - \varepsilon_u) \approx \sqrt{n \varepsilon_{PLC}} \quad \ldots \quad (5.2)
\]

where \( n \) is strain hardening exponent and \( \varepsilon_{PLC} \) represents the band strain at the point of maximum load.

Kang et al. (2005) also found a strong correlation between band strain and thickness reduction caused by the presence of PLC bands. As the band strain increases, the thickness reduction caused by PLC bands also increases. The higher the band strain, the greater the imperfection is in a sample. The greatest imperfection in a sample will potentially become the site of the diffuse neck. It can be concluded that a critical band strain is needed to produce the critical size of geometric imperfection which finally becomes the diffuse neck region.

The predicted necking strains from the model are then compared with those measured from uniaxial tensile results. The necking strains predicted by the model agree well with the experimental results in samples tested at room temperature with different strain rates, as shown in Figure 5.6. The predicted and measured necking strain in a sample with 0.02 of pre-strain still maintains good correlation. As the amount of pre-strain increases, however, a significant difference between predicted and measured necking strain starts to appear. A similar condition is also observed in aged pre-strained samples. This finding may be related to the fact that no geometric imperfection is formed, due to the suppression of the PLC effect, during pre-deformation at 223 K. Therefore, the
delay in achievement of critical band strain during room temperature test explains higher values of measured necking strains in comparison to model-predicted ones.

![Graph showing predicted and measured necking strain comparison](image)

Figure 5.6 Comparison of predicted and measured necking strain in tensile samples tested with different strain rates, different amount of pre-strain, and different aging time. Both aged samples were pre-strained up to true strain of 0.15 at 223 K.

5.5. Recovery Process and Band Strain in Aged Pre-strained Sample

A detailed observation of true stress-true strain curves of aged pre-strained sample (see Figure 4.8) suggests that recovery takes place during the aging period. The
mechanism of recovery is by dislocation rearrangement and dislocation annihilation
towards another structure of lower stored energy. It is well accepted that the flow stress,
\( \sigma \), scales linearly with square root of total dislocation density, \( \rho \), (Mecking and Kocks,
1981) as described in equation:

\[
\sigma = M\alpha G b \sqrt{\rho} \quad \ldots \quad (5.3)
\]

where \( M \) is the Taylor factor, \( G \) the shear modulus, \( b \) the Burgers vector, and \( \alpha \) is a
constant. Therefore, the effect of recovery on the mechanical response of deformed
material is reflected in the reduction of flow stress due to a decrease in the total
dislocation density. It is clear that the recovery process is a time dependent process, as
the reduction in flow stress increases with increasing aging time.

The recovery process in Al-Mg alloys has also been observed by Verdier et al.
(1999). They have studied the recovery of Al-2.5 wt.% Mg alloys in details by deforming
the material in rolling followed by heat treatment at temperature between 120\(^\circ\)C and
320\(^\circ\)C for durations varying between 1 minute and 1 week. The yield stress during
subsequent tensile test was measured in order to describe the kinetic evolution of
recovery process. They found that the decrease in yield stress is a logarithmic function of
recovery time at all temperatures. Based on TEM results, moreover, they suggested that
the recovery process appears to be independent of the evolution of dislocation
microstructure.

Moreover, an increase in the work hardening rate is also observed upon
subsequent deformation of aged pre-strained samples. Figure 5.7 shows a change in work
hardening rate, as manifested in a change in slope of true stress-true strain curve, as a result of the recovery process.

![Graph](image.png)

**Figure 5.7** An increase of work hardening rate, as manifested in a steeper slope of the curve, in aged pre-strained sample as a result of recovery process occurring during aging period. Both samples were pre-strained up to true strain of 0.15 at 223 K.

The occurrence of recovery also affects the development of band strain in aged pre-strained samples. Figure 5.8 shows a comparison of band strain between the non-aged and 120 hours aged samples, both of which were pre-strained up to true strain of 0.15 at 223 K. The band strain in 120 hours aged pre-strained sample starts off at much lower
value than that observed in pre-strained sample without aging. At a later stage, however, the band strain in both non-aged and aged samples falls down on the same trend-line. This finding corresponds with the flow behavior of aged pre-strained samples (see Figure 4.8), where a reduction in flow stress is observed only for the first four percent of plastic strain during reloading at room temperature. These observations suggest that the development of band strain is controlled by intrinsic material parameters, i.e. flow stress or work hardening rate.

Figure 5.8 A comparison of band strain in un-aged and aged pre-strained samples. Note that band strain in aged samples starts off much lower value.
This finding is in qualitative agreement with the model developed by Rizzi and Hähner (2004). Although their numerical simulations was performed specifically to capture kinematical characteristics of type A PLC bands, their model was built upon the same underlying principle governing the PLC effect, i.e. dynamic interactions between mobile dislocations and diffusing solute atoms. Therefore, it is considered valid to make a comparison between their simulation results with our findings.

Their model predicts a square root dependence of band speed, \( c_b \), on the hardening coefficient. During the propagation of a type A PLC band, the band strain, \( \Delta \varepsilon_b \), virtually accommodates the whole applied strain rate. Then, the following relation links the applied cross head velocity, \( v \), to the propagation speed \( c_b \):

\[
\Delta \varepsilon_b c_b = v \quad \text{..... (5.4)}
\]

Following equation 5.3, band strain is inversely proportional with band speed. Therefore, the faster the material hardens, the higher is the band speed, while the lower is the band strain. An increase of work hardening rate, which causes a decrease in band strain, observed in aged pre-strained sample are well agreed with the prediction of this model.

Band strain was found to proportionally increase with global strain during room temperature tests (presented in Figure 4.17). These results are in agreement with previous experimental observations in Cu-15 at.% Al. (Hähner et al., 2002 and Klose et al., 2003). Both models developed by McCormick and Ling (1995) and Rizzi and Hähner (2004) also predicted similar relationship between band strain and global strain. The number of dislocations involved in plastic deformation process increases as the global strain advances, which results in increasing band strain.
Although deformation band behavior is affected by prior deformation under positive strain rate sensitivity, band strain development in pre-strained samples shows little discernable difference with that observed in samples without pre-strain (see Figure 4.17 and 4.18). This suggests that band strain development is independent of the evolution of dislocation structure. However, extensive observations utilizing transmission electron microscopy (TEM) are needed to clarify the hypothesis.

5.6. The Impact of PLC Banding on Shear Banding

The development of a shear localization band in samples tested at room temperature with strain rates of $6 \times 10^{-4} \text{ s}^{-1}$ and $3 \times 10^{-3} \text{ s}^{-1}$ is described in Figure 5.9 (a) and (b). In both cases, there is only a single shear localization band leading to final fracture. This further confirms the observation of the result of shearing behavior presented in Figure 4.20, in which intense shear deformation occurred within one band.

The strain distribution in the shear band is highly non-uniform due to PLC banding, which occurs during the tensile test. By comparing both figures, it is found that the strain distribution within the shear band becomes more uniform as the applied strain rate increases. This seems logical as the band strain decreases with increasing strain rate, which causes smaller variations of plastic strain across the sample. Once the shear band is formed, further deformation is concentrated within the band itself. Finally, an increase in stress due to a decrease in the cross-sectional area of the sample overcomes the load carrying ability of the sample and leads to final fracture.
In contrast to those previous two samples, the sample with pre-strain of 0.235 shows uniform strain distribution within the shear band, as shown in Figure 5.10. This is due to a constant amount of plastic strain carried by PLC band, which shows Lüders front-like propagation behavior. The most important conclusion that can be drawn from
this observation is that there is no evidence of a relationship between PLC banding and shear banding. The formation of the shear band does not occur on the same plane, at which the PLC band propagates, instead it commences on a conjugate plane. This suggests that shear band formation in the 5754 Al-Mg alloy do not depend upon the occurrence of PLC banding.

Then, we can surmise that the development of a macroscopic shear band originates from the formation of grain-scale shear bands taking places in an early stage of deformation process. Indeed, Kang et al. (2006) have observed the development of microscopic slip lines, which are confined within individual grains, at small tensile strain in the same alloy. Extensive transmission electron microscopy of Al-4.8 wt.% Mg (Korbel et al., 1986) also showed the formation of micro-bands within the grains. As the
plastic deformation advances, these micro-bands are able to cross grain boundaries resulting in the development of a macroscopic shear band.

5.7. Shear Banding as a Failure Mode

It is shown in Figure 4.20 and Figure 4.21 that the 5754 Al-Mg alloy fails by shear band formation at both room temperature and 223 K. Once a macroscopic shear band forms, very little additional overall straining takes place while large strains accumulate in the band, leading to fracture. SEM observations of fracture surfaces also show characteristics of a shear mode of failure, i.e. shallow and elongated dimples. It is also known that the shear localization process in this alloy occurs prior to any damage accumulation (Spencer et al., 2002). All these findings suggest that plastic deformation localization enhances local damage processes and therefore plays a crucial role in determining the ultimate conditions for failure.

It is interesting to follow a sequence of shear localization processes in the sample that was pre-strained up to the necking point at 223 K and then reloaded until fracture at room temperature. This sample does not fail by shear localization within one band, as observed in all other samples, but instead by formation of a pair of conjugate shear bands, whose presence has been detected by both DIC and decoration techniques. The optical micrograph of a decorated fractured sample is shown in Figure 5.11.
By carefully examining DIC results, the way that this sample fails can be described as follows: the first shear band continues to work harden causing the microstructure within this band to become stronger than its surrounding microstructure. Further deformation is carried out by formation of a conjugate band which also shows some work hardening. This sequence of the development of conjugate shear bands was also observed in both α brass and copper (Weinrich and French, 1976). Finally, due to high plastic strain at the intersection of those two shear bands, a crack starts to form at this location and grows towards both sides of the samples.

It may be argued, however, that the formation of the second shear band is caused by the presence of a geometrical defect introduced by necking process. Therefore,
another experiment was performed using the same sample, in which the geometrical
defect after deformation at 223 K was removed by re-machining. A pair of conjugate
shear bands formation was still observed, although they are not formed at the same
location where the previous neck exists. It is evident that formation of the second shear
band is not due to geometrical reasons. Based on these results, it can be concluded that
the causes for localized shear plastic flow need not necessarily be linked to any subtle
microstructural softening effects. This statement agrees well with Asaro’s model (1979),
where shear band formation is possible with positive work hardening and without work-
softening condition. He suggested that localized shearing sets in when a critical value of
the work hardening rate over the flow stress is reached.
CHAPTER 6

SUMMARY

This work was focused on studies of two different phenomena, namely PLC banding and shear banding, occurring in 5754 Al-Mg alloy in order to characterize the interactions between these two phenomena. The main results are summarized as follows.

The 5754 Al-Mg alloy exhibited discontinuous yielding behavior, associated with the PLC effect, during uniaxial tensile testing at room temperature, at which the corresponding alloy exhibits negative strain rate sensitivity. The DIC technique has been used to indicate the appearance of type B PLC bands, characterized by random nucleation of deformation bands, at room temperature with a strain rate of $6 \times 10^{-4}$ s$^{-1}$. The PLC effect, however, was significantly suppressed at 223 K at the same strain rate, which consequently results in the higher ductility than that observed at room temperature. It can be concluded that the PLC effect causes a reduction in ductility by triggering an early onset of diffuse necking in uniaxial tension.

Random nucleation of deformation bands, associated with type B PLC bands, was observed during uniaxial tensile tests in both samples without pre-strain and samples pre-strained up to true strain of 0.02. Nonetheless, samples pre-strained up to true strain of 0.1 and 0.15 showed distinct propagative band behavior, instead of the random nucleation of deformation bands. This finding suggests that a critical amount of pre-strain is required
to ensure homogeneous distribution of solute atoms and dislocations, which in turn leads to a remarkable change in the nature of PLC banding observed in this study.

Furthermore, DIC measurements showed that band strain increases with increasing global strain in all cases. Prior deformation at 223 K has no influence on band strain as the results of band strain measurement in pre-strained samples followed the same trend line with those in samples without pre-strain. A considerable decrease in band strain, however, was observed in aged pre-strained samples as a result of the reduction in flow stress during reloading at room temperature. This observation supports the idea that band strain is controlled by internal material properties, i.e. flow stress or work hardening rate.

Finally, in a sample pre-strained up to true strain of 0.235 at 223 K, a final shear localization band was formed on a conjugate plane rather than at the same plane where PLC banding took place. This experimental finding suggests that there may be no correlation between PLC banding and shear banding.
APPENDIX

Following the analysis given by Marciniak et al. (2002), we consider a tensile strip in which a slight imperfection exists. The same load is transmitted by the uniform region and the imperfection. This can be written as:

\[ P = KA_0 e_1^n \exp(-\varepsilon_1) = K(A_0 + dA_0)(\varepsilon_1 + d\varepsilon_1)^n \exp[-(\varepsilon_1 + d\varepsilon_1)] \] .... (1)

Substituting \( \varepsilon_1 + d\varepsilon_1 = n \) for the strain at maximum load in the imperfection and \( \varepsilon_u \) for the strain in the uniform region, we obtain:

\[ \left( \frac{\varepsilon_u}{n} \right)^n \exp(n - \varepsilon_u) = 1 + \frac{dA_0}{A_0} \] .... (2)

Rewriting equation (2) gives:

\[ \left[ 1 - \frac{n - \varepsilon_u}{n} \right] \exp \left( \frac{n - \varepsilon_u}{n} \right) = \left( 1 + \frac{dA_0}{A_0} \right)^{\frac{1}{n}} \] .... (3)

In real cases, \( n - \varepsilon_u \) and \( \frac{dA_0}{A_0} \) are small compared with unity, therefore Equation (3) may be approximated as:

\[ \left( 1 - \frac{n - \varepsilon_u}{n} \right) \left( 1 + \frac{n - \varepsilon_u}{n} \right) = 1 - \left( \frac{n - \varepsilon_u}{n} \right)^2 \approx 1 + \frac{dA_0}{A_0} \frac{1}{n} \] .... (4)

Thus, we have:

\[ (n - \varepsilon_u) \approx \sqrt{-n \frac{dA_0}{A_0}} \] .... (5)
We now consider the PLC band as a geometric imperfection; thus the area change of $dA_0$ only relates to the band strain, $\varepsilon_{PLC}$. It is reasonable to assume that the width-strain is unchanged inside and outside a PLC band; then we obtain:

$$\frac{dA_0}{A_0} = \frac{W_{in}t_{in} - W_{out}t_{out}}{W_0t_0} \approx \frac{t_{in} - t_{out}}{t_0} \quad \ldots \quad (6)$$

subsequently:

$$\frac{dA_0}{A_0} = \frac{t_{in} - t_{out}}{t_0} = \varepsilon_{t,in} - \varepsilon_{t,out} = -\varepsilon_{PLC} \quad \ldots \quad (7)$$

where $W_{in}$, $W_{out}$, $t_{in}$, $t_{out}$, represent width and thickness inside and outside PLC band, respectively. $\varepsilon_{t,in}$ and $\varepsilon_{t,out}$ are the thickness strain outside and inside the PLC band, respectively, and $\varepsilon_{PLC}$ is the band strain.

By substituting equation (7) to equation (5), we have:

$$(n - \varepsilon_u) \approx \sqrt{-n \frac{dA}{A}} \approx \sqrt{n \varepsilon_{PLC}} \quad \ldots \quad (8)$$
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