

The Effect of Pre-Strain and Strain Path Changes On Ductile Fracture

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Abstract

Industrial metal forming operations generally require several deformation steps in order to create the final product. The mechanical behavior of materials undergoing strain path changes can be very different from those deformed in a given direction to fracture.

The work presented here employed laser drilled model materials to better understand the effect of pre-strains and strain path changes on void growth and linkage leading to fracture is studied.

The experimental results show that increasing pre-strain results in faster void growth which was justified in terms work hardening rate in the sample. Scanning electron microscope images revealed that the ductility of the sample decreased with increasing pre-strain but only slightly compared to the large decrease in far field strain at failure. This suggests that pre-strain affects strain localization significantly and to a lesser extent the ductility. Finally a finite element model has been built to predict the linkage between voids.

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Chapter 1

Introduction

Pre-straining is the accumulation of plastic deformation in a metallic component prior to its final use. Pre-straining can be deliberately and/or unintentionally imposed to parts used in industrial applications. This process is commonly followed by subsequent deformation steps, which can be in the same direction as the pre-strain or in a different one (also called strain path changes). There are many engineering scenarios where metallic components are subject to plastic deformation due to accidental loading, cold bending and ground movement. Therefore it is necessary to understand this process for different materials in order to predict the potential risks in industrial settings. Although the effect of pre-strain and strain path changes on different ductile fracture was well documented experimentally, there is no work so far on the prediction of void growth during plastic deformation for different pre-deformed metals.

This study uses laser drilled model materials to better understand the effect of pre-strain and strain path changes on ductile fracture in aluminum alloy 5052. By controlling the position, size and quantity of voids during material deformation, it is possible to visualize the ductile fracture process experimentally.

The preliminary objective of this thesis is to obtain quantitative information on void growth and coalescence for different pre-deformed material. Moreover, with extracted experimental results, the accuracy of the McClintock model for void growth and validation of finite element models can be assessed.

After a review of previous studies on ductile fracture in chapter 2, the model materials fabrication method is given in chapter 3 along with the experimental procedures used in the study. Chapter 4 presents and discusses the experimental results of the tests carried out on the laser drilled model material. Finite element modeling is presented in Chapter 5 and void growth is predicted with and without the damage approach. The main conclusions are presented in Chapter 6 and some suggestions are made for further research.

Chapter 2

Literature Review

2.1 Introduction

This section has three main parts. The first part of the literature review focuses on the physical processes of ductile fracture and the models, which are related to each fracture step. In the second and third parts of this section the effect of pre strain and strain path changes on ductile fracture will be discussed.

2.2 Physical ductile fracture process

Ductile fracture generally occurs in three stages. First voids are nucleated in the material (although it is possible to have some pre-existing voids as well). In the second stage, under high stress triaxiality, plastic deformation allows these voids to reach a critical void volume fraction. Finally when the voids are large enough and reach the critical value, they coalesce to form micro cracks. These may then grow into larger cracks that cause macroscopic and final failure (1). The current section describes the three stages of ductile fracture, namely void nucleation, void growth and void coalescence.

2.2.1 Void Nucleation

Discussion regarding void nucleation is more limited than the other stages of fracture. The reason lies in the difficulty in extracting material properties for both inclusions and inclusion/matrix interfaces which is essential to properly model the void nucleation process.

There are two mechanisms for void nucleation: inclusion failure and interface decohesion (or cavitation in the matrix). There are several models that predict void nucleation, but there are still some limitations to predict which of the two mechanisms will occur, stemming from difficulties in measuring material properties experimentally. Table 2-1 exhibits a summary of previous works on void nucleation and illustrates the parameters influencing void nucleation (1).

Table 2-1 Parameters influencing void nucleation (1)

Parameter (increase)	Inclusion failure strain	Inclusion/Matrix debonding strain	Authors
Inclusion size aspect ratio	↓	↓	Llorca et al. (2) Lee et al. (3) Brechet et al. (4)
Inclusion volume fraction	↓	↓	Babout et al. (5) Maire et al. (6)
Matrix strength	↓	=	Babout et al. (5)
Temperature	↑	↑	Zhao et al. (7)
Hydrostatic pressure	↑	↑=	French and Weinrich (8) Chen (9)

2.2.1.1 Void nucleation models

There are numerous approaches explaining void nucleation. The most important ones are analytical models made by Tanaka et al. (10) (energy criteria),

Argon et al. (11) (critical stress condition), Goods and Brown (12) (analytical particle radius), Thomason (13) (model based on stress), Kiser et al. (14) (mean field model based on Weibull statistics), and finite element models presented by Chu and Needleman (15), Needleman (16), Needleman and Tvergaard (17), Nutt and Needleman (18), Gonzalez and Llorca (19) and Segurado and Llorca (20).

Normally void nucleation requires the achievement of one of the following critical conditions: stress (for debonding) and strain (or energy level; for creation of a new surface at the inclusion/matrix interface).

There are a few approaches for predicting void nucleation mechanisms yet, much more work needs to be completed in this area, particularly with respect to obtaining local material properties.

2.2.2 Void Growth

Once a microvoid has been nucleated in a plastically deforming matrix, increasing plastic deformation will cause it to undergo a volumetric growth and shape changes essentially in the direction of the applied load. During the 1980s and 1990s, the fracture community has focused a lot of efforts on improving and extending void growth models.

2.2.2.1 Experimental results

To understand void growth, 2D approaches using metallographic techniques were carried out (21) (22) (23) (24) and it was shown how void growth is affected by plastic strain and stress state in the sample. However, these studies did not take into account volume fraction and shape of the voids. Marrini et al. (25) utilized

fractography technique to extract the volume of the voids in the metal matrix composite where the exponential increase in void growth with strain has been seen. Moreover, density measurement technique was used (26) (27) to measure void volume fraction and it was found that for a given plastic strain, the porosity increases by increasing inclusion volume fraction.

2.2.2.2 Void growth models

There are numerous studies related to void growth modeling, but most of them were modified from three basic model: McClintock (28), Rice & Tracy (29) and Gurson (30). Since, the Gurson model has lots of adjustable parameters, in the following section only the McClintock and Rice & Tracy will be presented.

2.2.2.2.1 McClintock model

The first effort in micromechanical modeling of ductile damage has been done by McClintock (28). His idea came from a solution for the deformation of elliptical holes in a viscous material and presents a criterion for ductile fracture by the growth and coalescence of cylindrical holes in a plastic material. In general McClintock uses power hardening relationship such as:

$$\sigma = A\epsilon^n \quad \text{Equation 2-1}$$

Where A and n are material constants. The major and minor axes of the initial hole are represented as $2b_0$ and $2a_0$, while 2b and 2a represent their instantaneous values as shown in Figure 2.1. He proposed this equation for modelling the growth of cylindrical voids:

$$\ln\left(\frac{R}{R_0}\right) = \epsilon \frac{\sqrt{3}}{2(1-n)} \sinh\left(\sqrt{3} \frac{(1-n)}{2} \frac{\sigma_1 + \sigma_2}{Y}\right) + \frac{\epsilon_1 + \epsilon_2}{2} \quad \text{Equation 2-2}$$

Where R is the mean radius of the elliptical void, R_0 is the initial void radius; ϵ is the far field true strain. The applied stresses σ_1 and σ_2 are those in the “a” and “b” directions; ϵ_1 and ϵ_2 are the corresponding strains.

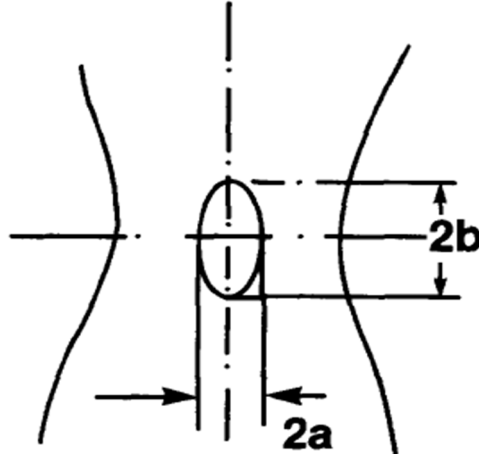


Figure 2-1: A schematic representation of the cross section of a tensile test specimen (28)

2.2.2.2 Rice and Tracy model

Rice and Tracy studied spherical voids instead of the cylindrical voids studied by McClintock. In both cases the results show that growth rates are significantly elevated by the superposition of hydrostatic tension on a remotely uniform plastic deformation field. In the case of a spherical void, Rice and Tracy (29) showed that the rate of variation of the void radius R can be exhibited for high stress triaxiality:

$$\left(\frac{R^0}{R}\right) = \alpha \exp\left(\frac{3}{2} \frac{\sigma_m}{\sigma_0}\right) \epsilon_{eq}^0 \quad \text{Equation 2-3}$$

Where α is a numerical factor, ϵ_{eq}^0 is the Von Mises equivalent strain and σ_0 the matrix yield stress. The initially proposed value for α (0.283) was modified by Huang (1991) to achieve a greater accuracy. In general the Rice and Tracy model

emphasizes the determining effect of the stress triaxiality, i.e. the ratio σ_m/Y , on void growth and has been successfully applied to a variety of materials and stress states.

However this model does not take into account the interaction between voids and the effect of void growth on material behaviour. This problem was solved by Gurson (30). He proposed a Von-Mises yield surface which is represented in a principle stress space, by a cylinder. If the state of stress is located inside the cylinder, the material deforms elastically while if it lies on the cylinder, the material deforms plastically.

2.2.3 Void Coalescence

Void coalescence is defined as a transition from a homogeneous void enlargement to a localized mode of plasticity between voids. There are three mechanisms for void coalescence as described below (31). The first mode of coalescence has been presented in Figure 2-2 (a), (c) and (e) is the internal necking mode of coalescence. The second one is shown in Figure 2-2 (b), (d) and (f) is the shear localization between large primary voids. This mode is frequently called “void sheeting”. The last mode of coalescence, which was proposed by Benzerga (32) (33), is less common and is called “necklace coalescence”, see Figure 2-2 (g) (31).

2.2.3.1 Void coalescence models

In general there are two common criteria for the initiation of void coalescence which are: maximum effective stress (Von Mises stress) and critical void volume fraction. In the first criterion, fracture initiates rapidly after the maximum effective stress has been reached and is valid for low stress triaxiality.

The second one indicates that void coalescence starts at a critical porosity; this is the most widely employed criterion in void coalescence modeling (31). Although there are numerous models that exhibit void coalescence, the majority of them are limited due to the lack of quantifiable experimental results. The main criteria of some models with their limitations are shown in

Table 2-2, which was presented by Weck (1).

In the following section the three main coalescence models are explained in detail. First, a geometrical model for onset of coalescence by Brown and Embury (34) is explained. Then, the Thomason (13) model (based on plastic limit load condition attainment) and Pardoen and Hutchinson (35) models are demonstrated for non-hardening and hardening material correspondingly.

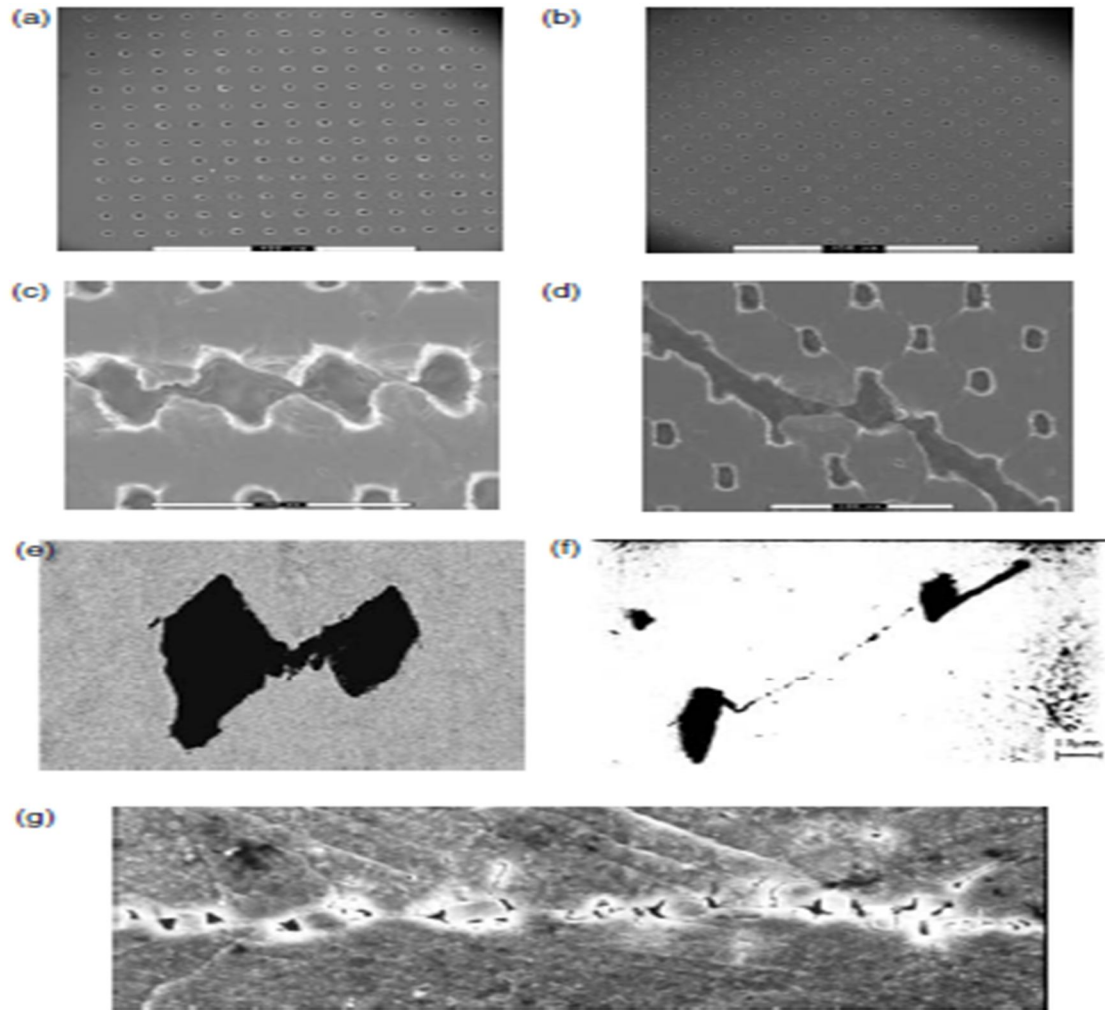


Figure 2-2: Micrographs showing the different void coalescence mechanisms in metals. The loading direction is vertical except for figure (g); from (a) to (d) the experiments made on perforated aluminum sheets (courtesy of A. Weck and D.S. Wilkinson, see also Weck et al., 2008) with different arrangements of voids: arrangement (a) leads to internal necking (see (c)) and arrangement (b) leads to coalescence in shear (see (d)); micrograph (e) shows an internal necking process in steel (source, Benzerga, 2000); micrograph (f) shows a void sheet mechanisms in steel with many secondary voids along the microshear band (source, Cox and Low, 1974); micrograph (g) is an example of void coalescence in column or 'necklace', seen in extruded copper bars (source, Pardoen, 1998). (31)

Table 2-2 Coalescence Models (1)

Author	Criteria	Limitations
McClintock(1968) (28)	Hole impingement	Cylindrical holes No interaction between holes No localization
Brown&Embury (1973) (34)	Void length equals intervoid spacing	For regular array of voids No Hydrostatic component No material properties
LeRoy (1981) (36)	Brown and Embury with stress triaxiality	No interaction between holes Regular array of holes No material properties
Tvergaard &Needleman (1984) (37)	Critical porosity and acceleration factor	Model rely on arbitrary parameters No hole geometry (average)
Thomason (1990) (13)	Plastic limit load	No microshear localization possible Only for non-hardening materials
Pardoen Hutchinson (2000) (35)	Plastic limit load with strain hardening	No microshear localization possible
Benzerga (2002) (32)	Plastic limit load for low triaxiality	No microshear localization possible
Gammage (2004) (38)	Stress equals global work hardening rate	Local work hardening = Global work hardening rate

2.2.3.1.1 Brown and Embury model

Brown and Embury (34) showed that coalescence occurs when shear bands at 45° can be drawn between the growing voids. This relates geometrically to the point at which the void length $2a$ is equal to the intervoid spacing $2W$, see Figure 2-3. This model shows an excellent agreement with the experimental results except for configuration at 45° and when the holes are close to each other (39). Lack of the state of hardening of the material experiencing coalescence (n) and effect of stress state are the main limitations of the Brown and Embury Model.

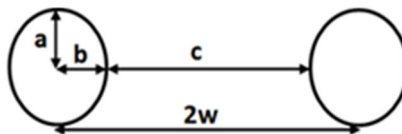


Figure 2-3: Schematic drawing showing of void dimension parameters for holes normal to the tensile axis

2.2.3.1.2 Thomason model

Thomason (13) proposed a new model for two dimensional and three dimensional plastic limit load of ductile fracture. His model is based on the attainment of a plastic limit load condition for localized plastic failure of the intervoid matrix. Figure 2.4 shows the geometry of this model. Thomason proposed that the onset of microvoid coalescence in 2D takes place when the following equation is verified:

$$\left(\frac{0.3A_{n-2D}}{a/c(1-A_{n-2D})} + 0.6 \right) (1 - V_f)^{-1} = \frac{\sigma_m}{Y} + \frac{1}{2} \quad \text{Equation 2-4}$$

Where A_{n-2D} is the area fraction of the intervoid matrix defined as $A_{n-2D} = e/W$, V_f is the initial void volume fraction defined as $V_f = (\pi/4) (c_0/W_0)^2$, σ_m is the mean stress and Y the plastic equivalent stress. For 3D analysis Thomason proposed a similar equation:

$$\frac{0.1}{(a/c)^2} + \frac{1.2}{\left(\frac{c}{c+e}\right)^{1/2}} (1 - V_f)^{-1} \left(1 - \left(\frac{3\sqrt{\pi}}{4} V_f \right)^{2/3} \left(\frac{b}{b_0} \right)^2 e^{\varepsilon_1} \right) = \frac{\sigma_m}{Y} + \frac{3}{2} \quad \text{Equation 2-5}$$

As it has been shown above the Thomason (1990) model works for both 2D and 3D void coalescence but it is valid only for non-hardening materials. This gap in the model motivated Pardoen and Hutchinson (35) to modify Thomason model.

2.2.3.1.3 Pardoen and Hutchinson model

Pardoen and Hutchinson (35) used cell model computation for the axisymmetric growth and coalescence where accurate predictions have been seen for a wide range of porosity, strain hardening, stress triaxiality, initial void shape and void spacing in elastoplastic solid. The Pardoen and Hutchinson model considered the limitations of previous models and proposed some possible

improvements. The constant values of 0.1 and 1.2 in equation 2-5 for 3D Thomason are replaced by two parameters, α and β that incorporate a dependence on the strain hardening exponent n where β is almost constant and can be taken as 1.24 while

$$\alpha(n) = 0.1 + 0.217n + 4.83n^2 \quad (0 \leq n \leq 0.3) \quad \text{Equation 2-6}$$

By implementing this change the Thomason model can be used for hardening materials as well. In Figure 2.5 the variation of α and β as a function of the strain hardening exponent is shown.

The coalescence modeling has been described in this section by various methods but due to the lack of experimental data, the effect of the parameters dictating coalescence has not been validated experimentally.

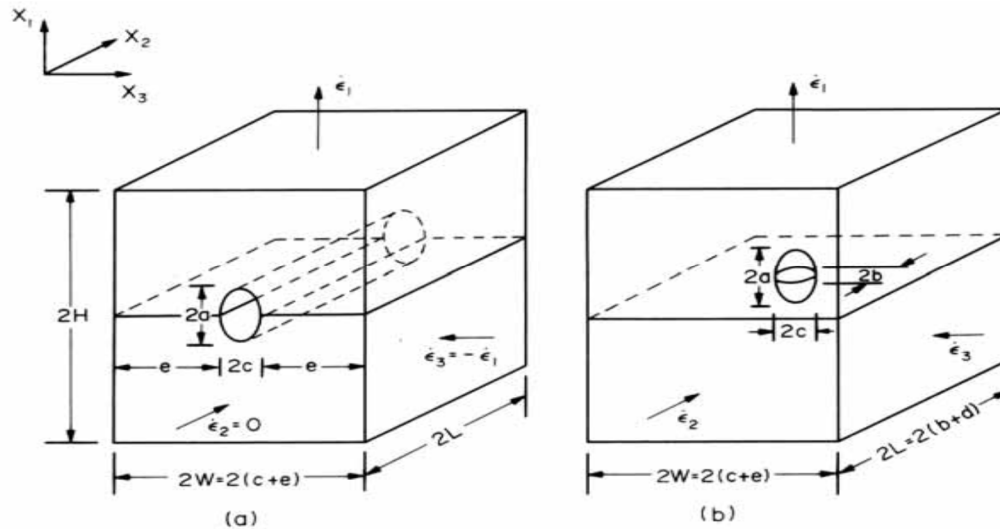


Figure 2-4: The unit cells of similar planar geometry for (a) two-dimensional models with prismatic elliptical voids and (b) three-dimensional models with ellipsoidal voids. (13)

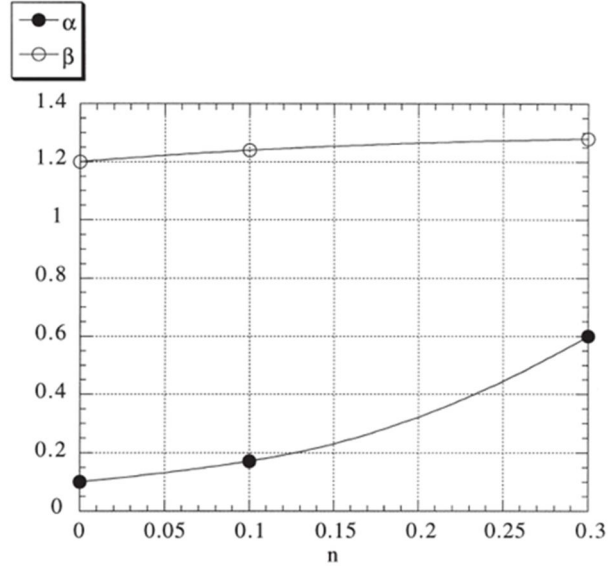


Figure 2-5: Variation of the parameters of the coalescence model, α and β as a function of the strain hardening exponent. (35)

2.3 Effect of pre-strain on material properties

2.3.1 Pre-Strain definition

Figure 2-6 presents the tensile pre-strain and compressive pre-strain histories that may exist in engineering applications. The terminology tensile pre-strain refers to applying a load in tension on the material, unloading it, followed by a final tensile test to failure. The final tensile test and pre-strain history are applied in the same direction. Compressive pre-strain means that the specimen is first compressively loaded and unloaded, followed by a tensile test to failure. In the latter case the direction of compressive pre-strain is opposite to the direction of final tensile test, see **Figure 2-6**. According to the goal of the current research, the remainder of this review will be focused on tensile pre-strain history and verifies its effect on different mechanical properties of materials, particularly the aluminum alloy 5052.

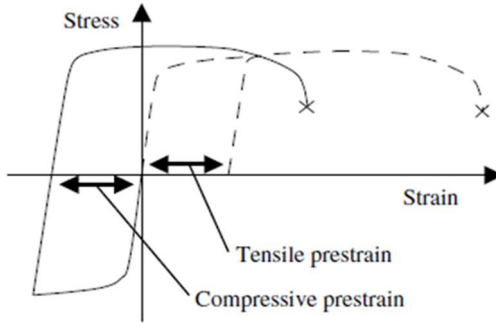


Figure 2-6: The definition of compressive and tensile pre-strain. (40)

2.3.2 The importance and application of pre-strain

The pre-straining process is a consequence of many industrial processes. There are many engineering scenarios where metallic components are subjected to plastic deformation (pre-strain) due to accidental loading, cold bending and ground movement to name a few. Pre-strain can have a negative effect on fracture and it is therefore necessary to understand this process to predict the potential risks in industrial settings. In the following examples, the parameters and material properties affected by pre-straining will be discussed.

During fabrication of naval structures, plates of HSLA steel are cold deformed into shape and welded together. Cold deformation is essentially a pre-straining process that changes the dislocation structure of a material. It is expected that cold deformation has a large influence on the fracture behaviour of a material (41).

In the case of an earthquake, energy is expected to be absorbed through plastic deformation of structures to prevent failure. Appropriate plastic deformation is required for building security, making investigations into the effect of plastic deformation on fracture toughness of structural steels important (42).

One of the main applications for pre-strain history is in the automobile industry. For example factories use interstitial free (IF) steel for fabricating automobile components such as panels. The forming of these panels often introduces a significant amount of pre-strain in the IF steel. Thus it is necessary to know the fracture behaviour of these materials after pre-straining to ensure good formability (43).

Another example of pre-strain history is the pipe reeling process in offshore drilling. The reeling process is an efficient and economical way to install pipelines on the seabed. During the process plastic deformation as large as 2-3 % can occur in the pipe, so the distribution of the pre-strain history over the pipe's cross section should be predicted precisely. The effect of pre-strain history on subsequent deformation and fracture behaviour must be studied for safety reasons (44).

Moreover, in many metal forming and removal processes, such as stretching and deep drawing of plates, cold-punching of the holes and guillotining of thin sheets, deformation is introduced. Therefore more experimental data for pre-strain effect on fracture behaviour of thin metal sheets is needed (45).

It has been shown that there are different scenarios where pre-strain occurs in industrial materials. In the following sections the loading and deformation capacity of components will be discussed for predicting the failure in materials and increasing the safety factor in industrial settings.

2.3.3 Effect of pre-strain on microstructure evolution

2.3.3.1 Effect of pre-strain on microstructure

D. Wang et al. (46) used a transmission electron microscope (TEM) to take images of the 7050 aluminum alloy with (5% pre-strain) and without pre-strain history. In the sample without pre-strain there are homogeneous precipitate particles, but in the 5% pre-strain sample some large η -phase precipitates were observed within the grains and there were no other precipitates around these large precipitates, see Figure 2-7. In fact, the dislocation creates the efficient sites for precipitate nucleation. These large precipitates formed after pre-straining the sample could lead to early material damage. This research shows the importance of pre-strain on microstructural and thus mechanical changes in materials.

Jong-hyun Baeka et al. (47) who worked on the mechanical properties of API 5L X65 pipe showed that microstructures without pre-strain and with pre-strain are not the same. As it is shown in Figure 2-8, by increasing the pre-strain up to 10%, the pearlite microstructure is extended due to the axial tensile pre-strain. It was found that this increased volume fraction of pearlite is due to the creation of a cementite network around the previously-austenitic grain boundaries.

2.3.3.2 Effect of pre-strain on dislocation activity

It is well documented that dislocations will be generated during plastic deformation in metallic material (46). It was found that the dislocation density after the 5% pre-strain is high within grains and around grain boundaries. Dislocations also provide some effective sites for precipitate nucleation and in the aging process they act as diffusion paths for precipitation elements (48), (49). Dislocations can

aggregate around the grain boundaries and provide heterogeneous nucleation site for η -phase particles in API 5L X65 specimen, so around grain boundaries the precipitates were smaller.

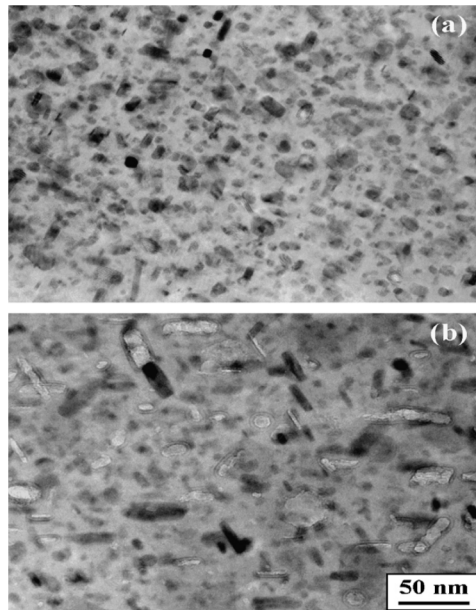


Figure 2-7: TEM micrographs showing precipitates within grains of 7050 aluminum alloy sample: a) without pre-strain, b) 5% pre-strain (46)

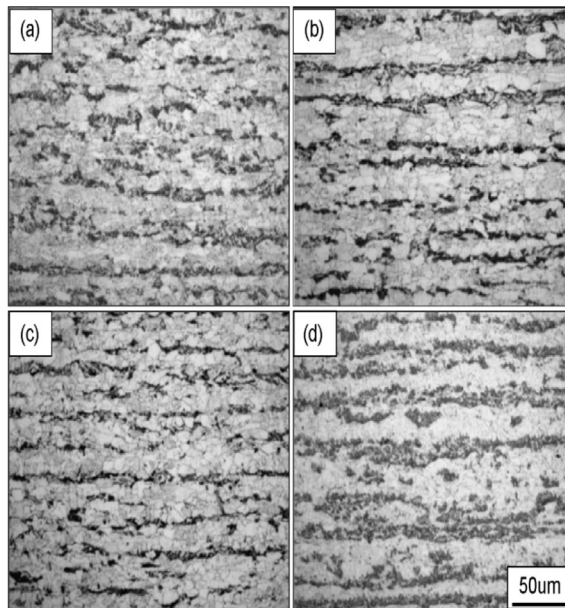


Figure 2-8: Microstructures for virgin and pre-strained materials: (a) virgin material, (b) 1.5% pre-strain, (c) 5% pre-strain, and (d) 10% pre-strain (47).

2.3.4 Effect of prestrain on the mechanical properties

2.3.4.1 Effect of pre-strain on tensile properties

The mechanical properties of API 5L X65 pipe were studied by Jong-hyun Baek et al. (47). They applied pre-strains of 1.5, 5 and 10% to the plate-type tensile specimens. They conducted tensile tests on the pre-strained specimens, and plotted the results for different parameters; see Figure 2-9.

From Figure 2-9, it can be seen that the elastic region has the same behaviour and there is no significant change between pre-strained and 0% pre-strain samples. The slope of the stress-strain curve (after elastic part) decreases with increasing the pre-strain. Due to pre-strain applied prior to the tensile tests, strain to failure decreases by increasing the pre-strain. However, increasing pre-strain up to 10% also causes a significant increase in yield strength as the material hardens.

2.3.4.2 Fracture toughness

The effect of pre-strain on fracture toughness has been studied for different types of pre-straining and materials. S. Sivaprasad et al. (40) have worked in this area and made a brief overview of this investigation, which is shown in Table 2-3. From Table 2-3, it is clear that the existing results on the effect of pre-strain on fracture toughness do not show any generalized pattern. For more detailed analysis S. Sivaprasad et al. (40) used two varieties of Cu-strengthened HSLA steel for their experiments. They imposed 1, 2, 3, 4 and 5% pre-strain on the specimens during a tensile test and used the J-R curve to show changes in J_c (critical fracture toughness) with respect to the amount of pre-strain, see Figure 2.10. In both steels, it can be seen that J_c increased slowly, up to ~2% pre-strain, and then decreased

deleteriously. This initial increase in fracture toughness up to 2% pre-strain is significant since, from the viewpoint of designing with the grades of HSLA steels being studied, it shows the extent of cold deformation that can be applied to steel plates without any loss in critical fracture toughness, J_c . This variation of J_c with pre-strain was clarified in terms of the variation of the tensile flow properties of the steels with pre-strain.

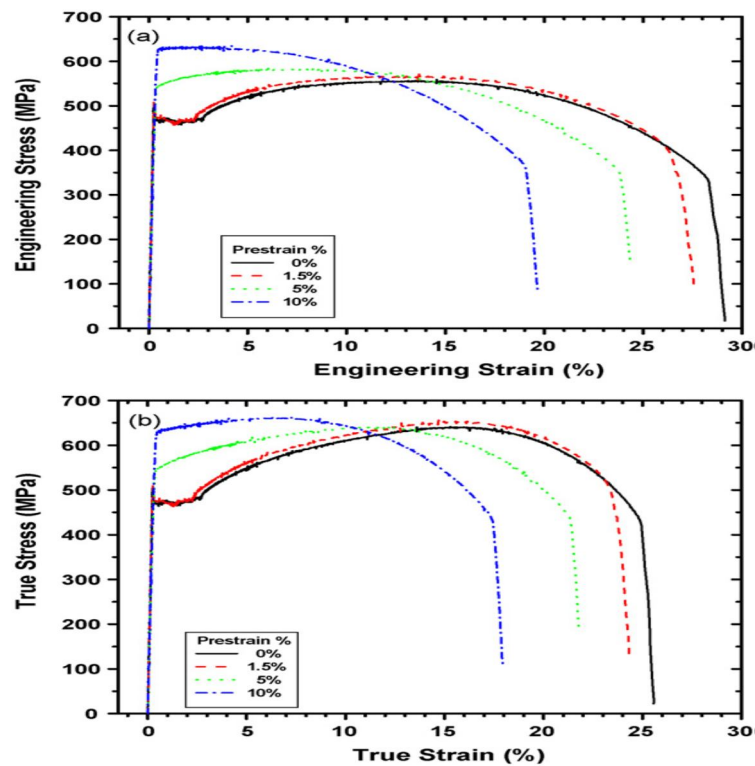


Figure 2-9: Stress-strain curves for API X65 pipes with pre-strain: (a) engineering stress-strain curves and (b) true stress-strain curves (47)

Table 2-3: Summary of pre-strain effects on fracture toughness (41)

Nature of Pre straining	Material	Reported effects	Reference
Tensile and cold rolling	AISI 4340 (annealed) 310 SS	J_{IC} increases with prestrain for annealed 4340 steel. - For 310 SS, J_{IC} decreases with prestrain. - Nature of variation of fracture toughness is the same for both modes of prestrain, however, the magnitude varied	(50) (51) (52)
Tensile and cyclic pre straining	AISI 4340 (Q&T) 316 SS	J_{IC} or J_M decreases with tensile prestrain for both the steels. - For quenched and tempered 4340 J_{IC} or J_M increases with cyclic prestrain. - For 316 SS, J_{IC} or J_M decreases with cyclic prestrain.	(53)
Compressive	HY-80	No significant change in CTOD of K_{IC} for prestrain up to 0.17, followed by a decrease.	(54)
Tensile	Al-alloy	K_{IC} decreases with prestrain.	(55)

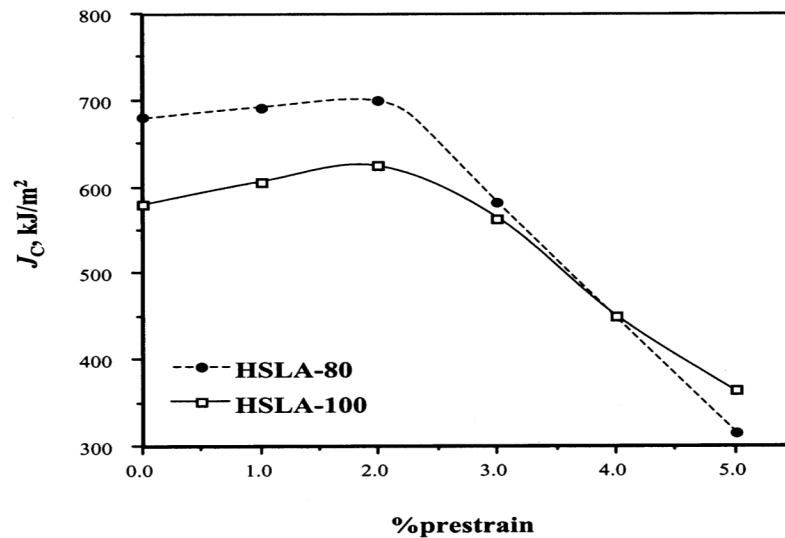


Figure 2-10: Effect of pre-strain on ductile fracture toughness of HSLA steels (41).

Table 2-4: Mechanical properties (42)

	SN490	PS-SN	X65	PS-X	SA440	PS-SA
σ_{ys} (MPa)	348	530	350	631	469	665
ϵ_f	1.35	1.21	1.71	1.55	1.5	1.33

ϵ_f : strain to fracture for smooth axisymmetric tensile specimens.

2.3.4.3 Yield strength (YS)

The effect of pre-strain on the yield strength (YS) of the HSLA steels (80 & 100) is presented in (41). For HSLA-80 steel it can be seen from Figure 2-11 that the yield strength does not change up to ~2% pre-strain. This constancy is related to the ~2% yield propagation exhibited by the HSLA-80, which displays a mechanism that prevents hardening on pre-straining. In contrast the HSLA-100 steel's YS increases with pre-strain. This behaviour is additional proof of the fact that the HSLA-100 steel does not follow from yield propagation effects. Also the effect of pre-strain on ultimate tensile strength (UTS) was found to be similar to the behaviour of yield strength.

2.3.4.4 Hardening exponent component, n

For investigating the effect of pre-strain on the strain hardening exponent, n, S. Sivaprasad et al. (41) used the Hollomon equation: $\sigma = k\epsilon^n$ for their experiment, under the assumption that true uniform strain, ϵ_u , is numerically equivalent to n. They modified ϵ_u and defined a new true strain, ϵ_{u^*} , to exclude elastic strain.

The effect of pre-strain on true uniform strain, ϵ_{u^*} (or hardening exponent component, n) is presented in Figure 2-12. It can be seen from the Figure 2-12, ϵ_{u^*} for both HSLA-80 and 100 does not change up to ~2% pre-strain, after which it decreases by approximately the amount of additional pre-strain applied. This variation is similar to the change in critical fracture toughness, J_c , with pre-strain, which was shown in Figure 2-10.

2.3.4.5 Model to correlate J_c & tensile parameters

The similar behaviour of tensile parameters and fracture toughness with pre-strain encouraged S. Sivaprasad et al. (41) to propose a model for predicting fracture toughness from tensile parameters of ductile materials. Initially they used the model proposed by Hahn and Rosenfield (56) (fracture initiation from crack tip), and determined that this model is in good agreement with experimental results for low toughness materials. For high toughness materials it has a deviation up to $\pm 30\%$, which was not practical. Finally they modified the previous model and derived this relation:

$$J_c = 2 C m (\epsilon_u^*)^2 \epsilon_f \sigma_0 \quad \text{Equation 2-7}$$

Where m is a constant with a value between 1 and 2 (57), σ_0 is the flow stress, C is a constant equal to 0.025 (58) and ϵ_f is the critical crack tip strain.

Finally they found that critical fracture toughness values, J_c , derived from the model is similar to the experimental values but with some deviations which is due to Hahn and Rosenfield's (56) assumption. They also used initiation fracture toughness J_i from the experimental J-R curve for better comparison and compared it with modeling results. Both experimental and model values have the same trend with pre-strain, especially for low pre-strain levels where the model results are within $\pm 5\%$ of the experimental results, see Figure 2-13.

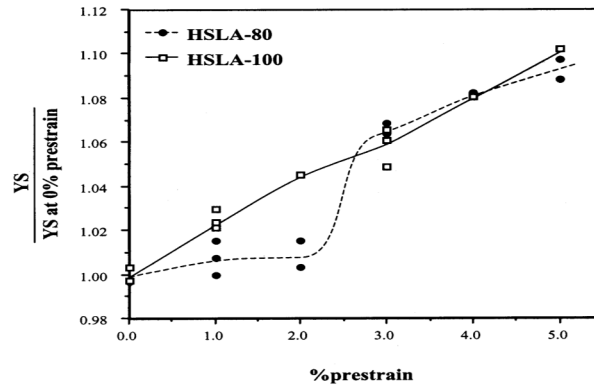


Figure 2-11: Effect of pre-strain on yield strength of HSLA steels; relative effects (41)

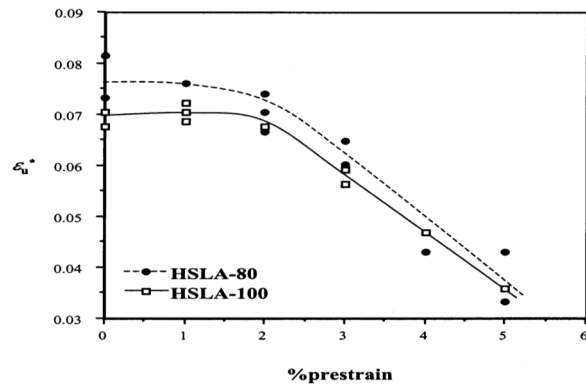


Figure 2-12: Variation of true uniform strain ϵ_{u^*} with pre-strain for HSLA steels (41)

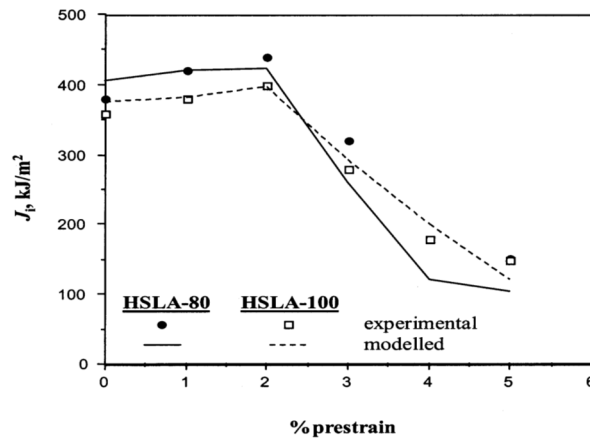


Figure 2-13: Variation of initiation fracture toughness of HSLA steels with pre-strain; both experimental and model predictions are shown (41)

2.3.5 Effect of pre-strain on ductile fracture initiation

Keitrao Enami (40) studied the effect of both compressive and tensile pre-strain on ductile fracture initiation in steels and compared it with the numerical method as follows. He used Hancock et al. (59), definition for the initiation of ductile fracture: when the maximum tensile load point is reached in the load-displacement curve, the load starts to decrease gradually. During this load decrease a sudden sharp change of the downward slope occurs in the curve, which is defined as 'point S' (the point of ductile fracture initiation), see Figure 2-14.

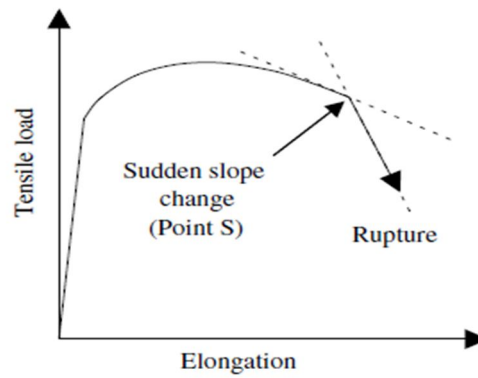


Figure 2-14: Schematic illustration of the load–elongation curve in the round notched tension test (40)

In addition to above definition Keitrao Enami (40) suggested the following equation for the relationship between stress triaxiality and critical equivalent plastic strain, where α and β are material constants.

$$\varepsilon_f = \alpha \exp\left(\frac{-\beta \sigma_m}{\sigma_e}\right) \quad \text{Equation 2-2-8}$$

Keitrao Enami (40) did pre-strain tests for two different steels. After pre-straining and machining (round notching) the specimens, tensile tests to failure were conducted at ambient temperature. Figure 2-15 (a)-(d) represents the relationship between tensile load and gage elongation for steel-A specimen in tension. From

Figure 2-15, it can be seen that the steel A specimens failed by ductile fracture in 0%, 10%, 30% compressive and 10% tensile pre-strain. And in the cases of 10%-1R and 30%-1R compressive pre-strain for steel B local cleavage cracks were observed.

Also they found that the maximum tensile load increases with decreasing notch radius; it means that the ductile fracture initiation point, S, decreases when the notch radius decreases. By comparing the experimental results with the FEM (axisymmetric model for rounded notched specimen) it can be seen that both values are in good agreement and completely consistent with each other.

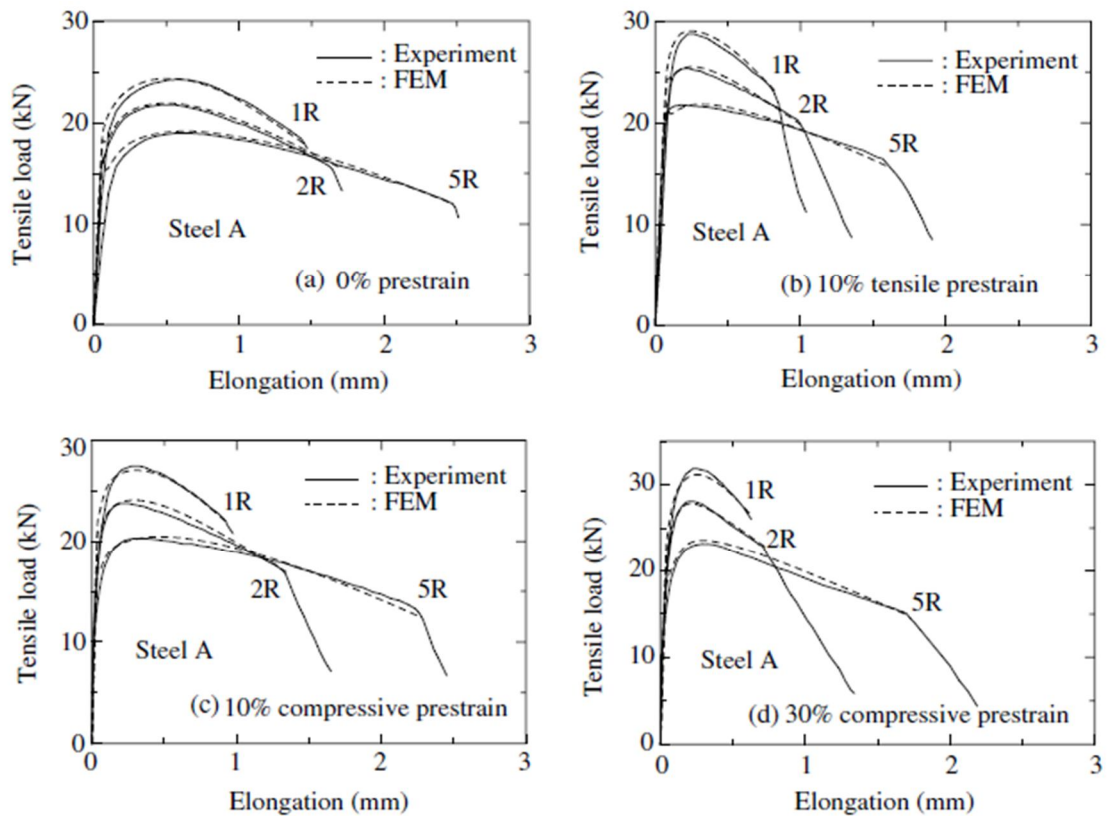


Figure 2-15: Relationship between tensile load and gage elongation in the round notched tension test for Steel A: (a) 0% prestrain; (b) 10% tensile prestrain; (c) 10% compressive prestrain; (d) 30% compressive prestrain (40)

2.3.5.1 Effect of pre-strain on void density

The effect of tensile pre-straining on the ductile fracture behavior of Nb-Ti stabilized IF (Interstitial-Free) steel has been verified with regards to the void nucleation strain by Sharabani Majumdar and K.K. Ray (43).

From the X-RAY (EDX) analysis it was shown that the void density increases with the radial strain ($\ln(A_0/A_f)$) percentage, see Figure 2-16. From this figure it can be seen that there are two regimes of void density variation, which are attributed to the separation of inclusions (regime I) and precipitate particles (regime II) from the matrix. So, void nucleation occurs by inclusions at lower strain and by precipitates at higher strain.

This observation can be prescribed to the possibility that non-metallic inclusions size is usually narrow compared to precipitate particles. The critical strain for void nucleation (ϵ_n) for the precipitate particles has been studied for the number of pre-strained tensile specimens.

The value of ϵ_n first rises to a critical value of pre-strain which can be attributed to the decrease in the value of n_1 ($n_1=1-n$; n =Strain hardening exponent) and then suddenly drops with further increase of pre-strain which can be explained with Brown and Stobb's dislocation model (60), see Figure 2-17. In fact, pre-strain increases the local stress due to the dislocation density around the particles. This model indicates that void nucleation starts at the interface of hard particles and ductile matrix when the result of three different stress contributions surpasses the particle-matrix interfacial strength; (1) high local stress at the PMI, (2) total flow stress around the particles, and (3) superimposed hydrostatic stress, (60).

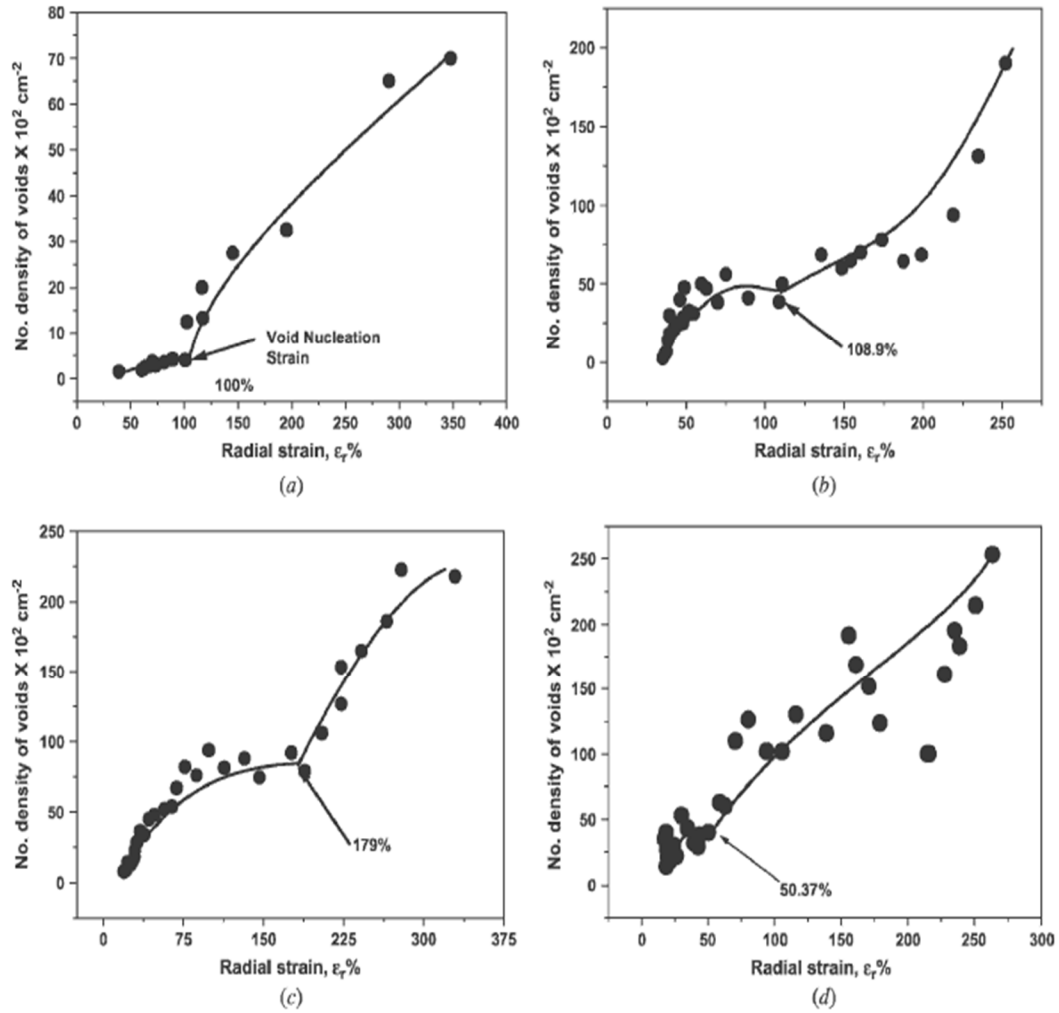


Figure 2-16: The variation of the number density of voids with radial strain in the fractured specimens of investigated IF steel with varying amounts of prestrain: (a) 0 %, (b) 4 %, (c) 6.7 %, and (d) 12 % (43).

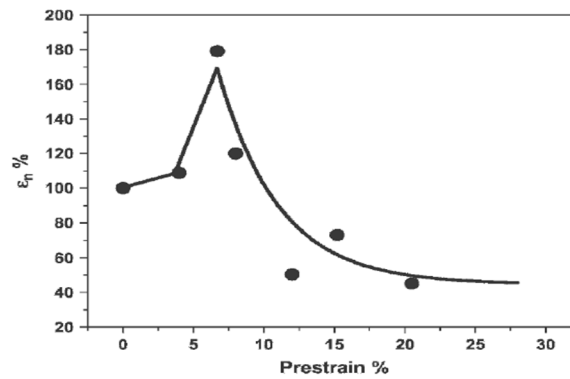


Figure 2-17: The variation of critical strain for void nucleation (ϵ_n) with pre-strain in specimens of the investigated IF steel (43).

2.3.5.2 The effect of pre-strain on dimple size

Sharabani Majumdar and K.K. Ray (43) showed that the variation of the dimple size with pre-strain for void nucleation has the same trend as that of critical strain with pre-strain, see Figure 2-18. They used Brown and Stobb's dislocation model (60) to explain this decrease after 6.7% pre-strain: when the distance between the two void nuclei decreases, the adjacent voids has less space to grow. Hence the void/dimple size drops with the increase in pre-strain after 6.7.

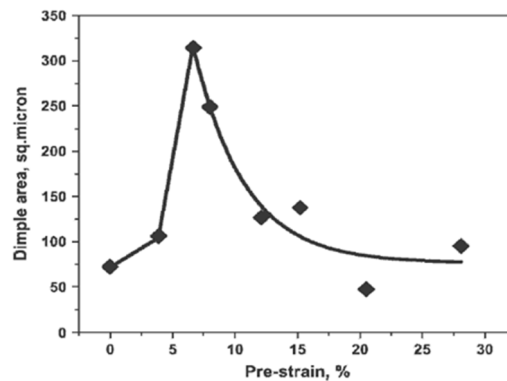


Figure 2-18: The variation of dimple area on the fractured surfaces with the amount of pre-strain in the specimens (43).

2.3.6 Effect of pre-strain levels on void growth, modeling

The variation of void shape with respect to pre-strain history was analyzed by B. Skallerud et al. (61). They intended to use a model in load-controlled manner; for this purpose the finite element software ABAQUS was utilized. Figure 2-19 represents the effect of pre-strain on void shape changes at a constant stress triaxiality of 1.33. In the case with 0% pre-strain the void aspect ratio, S , increases with plastic deformation and tends towards a prolate shape. However for the void with 10% pre-strain the void aspect ratio decreases with increasing plastic

deformation, and for 5% pre-straining the void aspect ratio first increased up to 1.18 after which it will be constant with further plastic deformation. It is also interesting to note that the final void aspect ratio is independent from pre-strain history at constant stress triaxiality. It means that at the onset of void coalescence the void aspect ratio is the same for all the pre-strain scenarios; see solid points in Figure 2-19.

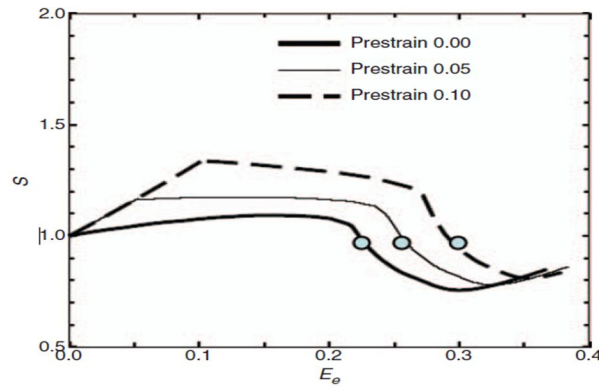


Figure 2-19: Effect of pre-strain levels on void shape change for $f_0=1.04\%$ and $T=1.33$. In the figure the solid point indicates the starting point of void coalescence (61).

2.3.7 Effect of pre-strain on near crack tip stress fields, modeling

There are several factors that affect the fracture toughness and drive the fracture process. Stress and strain near the crack tip in the specimen play a significant role in fracture steps. Prediction of the effect of pre-straining on the crack tip stress field was demonstrated by G. C. Xia et al. (62). They used FEM method with ABAQUS for the CTOD (crack tip opening displacement) specimen. The variation of maximum principal stress σ_1 near the crack tip is shown in Figure 2-20. From Figure 2-20, it is found that the stress fields near the crack tip in CTOD specimen increase by increasing the pre-strain history due to an increase in flow stress by 2% pre-

straining. Also this elevated crack tip stress fields promotes brittle fracture of the CTOD specimen.

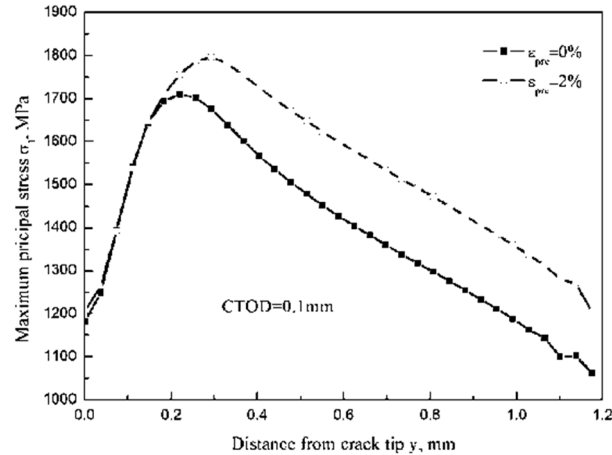


Figure 2-20: Effect of pre-strain on stress fields near crack tip of CTOD specimen (62)

2.3.8 Effect of pre-strain on void coalescence strain, Modeling

The effect of pre-strain on void coalescence with and without pre-straining was studied by B. Skallerud et al. (61). They used the ABAQUS RIKS method for model material where the nodal constraints were applied in CTOD specimen. The initial void volume fraction has been taken as 1.04%. In this study the model material used different stresses triaxiality values and different initial void geometries. The general observation was that the stress triaxiality and void aspect ratio have a significant effect on void coalescence strain in comparison to pre-strain history, which has less effect especially for a prolate void shape; see Figure 2-21.

It was found that during the void coalescence step the maximum axial stress position in the void ligament is an indicator for void coalescence. With increasing plastic deformation the position of this indicator shifts from the void edge to cell boundary and coalescence takes place when the position appears at the boundary.

So, the void coalescence strain decreases with increasing the stress triaxiality and pre-strain history. However the effect of pre-strain on void coalescence strain is less for prolate initial void shape; see Figure 2-21 (c), in comparison to spherical and oblate ones. For the case of a stress triaxiality less than 1.33 for prolate voids, this effect is negligible but pre-strain history causes a strong reduction in void coalescence strain in the case of oblate void shape, see Figure 2-21 (b).

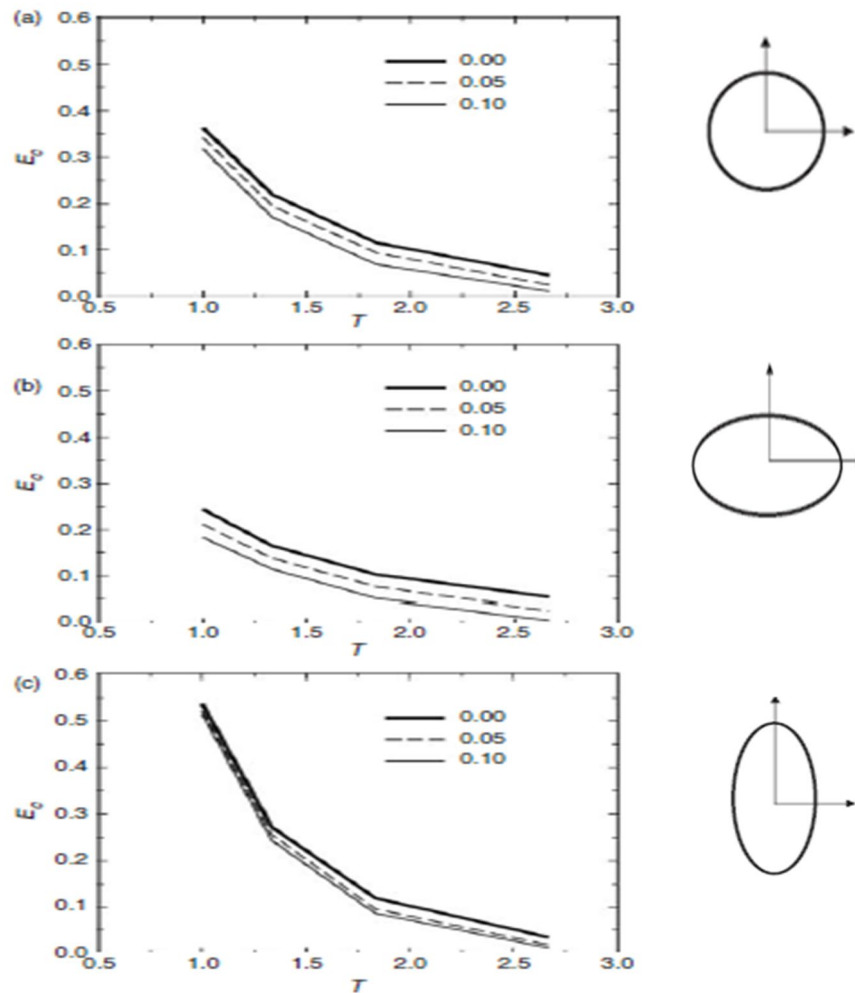


Figure 2-21: Effect of pre-strain on the void coalescence strain, (a) spherical void with $S=1$, (b) oblate void with $S=0.25$, and (c) prolate void with $S=4$. The numbers in the legend indicate the pre-strain levels (61).

The primary purpose of this literature review was to study previous works on pure pre-strain (without strain path effect), which plays an important role in affecting ductile fracture. Although there are numerous parameters that have a strong effect on ductile fracture, this study attempted to focus on the most important ones. Table 2-5 represents the effect of pre-strain on different parameters during ductile fracture process.

Table 2-5: Effect of pre-strain on different parameters during ductile fracture process

Parameter	Effect of pre-strain	Reference
microstructure evolution	- In the sample without pre-strain there are homogeneous precipitate particles, but in the 5% pre-strain sample some large η-phase precipitates were observed within the grains - dislocation will be generated during plastic deformation in metallic material (using TEM)	(46)
mechanical properties of metals	- Hardening became less with more prestrain. - Strain to fracture decreases by prestraining. - Increasing prestrain up to 10% cause to increase in yield strength - YS increase with prestraining - For K_{IC}, see table 2.3 "n" does not change up to ~2% prestrain, and then it decrease	(47) (41) (56) (42)
ductile fracture initiation	- Maximum tensile load increases with decreasing notch radius; it means that the ductile fracture initiation point, S, decreases when the notch radius decreases. - And in the cases of 10%-1R and 30%-1R compressive prestrain for steel B local cleavage cracks were observed - The value of ϵ_n first rises to a critical value of prestrain and then suddenly drops with further increase in prestrain - The variation of the dimple size with prestrain for void nucleation has the same trend as that of critical strain with prestrain	(40) (59) (63) (43) (64)
void growth & crack tip stress fields	- Generally, ductile fracture initiation point starts to decrease by increasing prestrain - stress fields near the crack tip in CTOD specimen increase by increasing the prestrain	(61) (62)
void coalescence strain	The void coalescence strain decreases with increasing the stress triaxiality and prestrain history. However, the effect of prestrain on void coalescence strain is less for prolate initial void shape	(61)

2.4 Effect of strain path changes on material properties

2.4.1 Strain path changes; Definition and Application

During metal forming processes, multi-stage forming operations are carried out, which means that the material is strained in different directions or complex strain paths (also called strain path changes) in order to obtain the final shape.

For instance in automotive industries more than 500 parts are manufactured by these multi-stage forming operations (65). In packaging industries, any change in the strain path may lead to significant changes in microstructure, deformation texture, mechanical properties and failure of the material. Predicting the formability and the strength of parts undergoing such multi-stage forming operations is difficult. This means that costly conservative and trial-and-error approaches have to be taken by manufacturing industries. In the following section, the complex strain path and its effect on different material properties will be discussed.

2.4.2 The effect of strain path changes on microstructure and texture evolution

The texture and microstructure evolutions in IF-Ti steel during complex shear-strain paths were studied by E.V. Nesterova et al. (66). The observed types of microstructure in grains of various orientations depend on two parameters: orientation path and deformation path. According to this concept these texture and microstructure evolutions were classified into seven categories:

1. Type A: formation of well-developed dislocation walls parallel to the shearing direction, Figure 2-22.
2. Type A*: partial dissolution, Figure 2-23.
3. Type B: reorganization of the dislocation boundaries.

4. Type C: formation of well-developed dislocations perpendicular to the shearing direction.
5. Type D: fragmentation case.
6. Type E: formation of two sets of dislocations boundaries and forms in grains in which the activation of two slip planes is expected.
7. Type F: associated with microband formation.

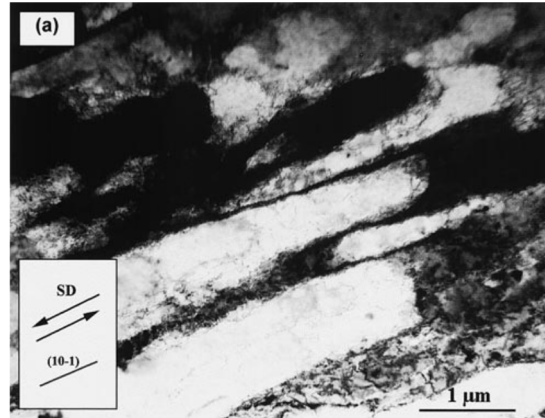


Figure 2-22: TEM microstructures of γ -fiber grains after 15 and 60 % of monotonic amount of shear along RD: (a) 15 %, γ_1 grain, (ND) [RD] = (0.57, 0.45, -0.69) [0.30, -0.89, -0.34] (66).

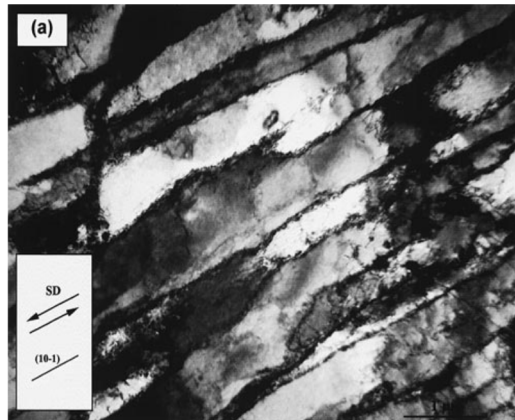


Figure 2-23: TEM microstructures of γ -fiber grains after Bauschinger shear tests (30 % shear in the forward direction and 30 % in the reverse one): (a) γ_1 grain, (ND)[RD] = (0.66, -0.49, -0.56)[0.39, 0.41, 0.82] (66).

It was concluded that type A will happen in the monotonic tests, type A* and type D will predominate after Bauschinger sequences (tension-compression) resulting in a decrease of the work-hardening rate, and type F will be observed

during orthogonal sequences which produces the softening effect on the hardening curves.

Another approach was presented for studying the effect of strain path changes on microstructure and texture evolution during two stages strain paths, specifically on dislocation sheets in an IF steel, B. Peeters et al. (67). The dislocations clusters in some grains of an IF steel were captured under a TEM and it validated the extended Taylor model predictions where the polarity and intensity of dislocation sheet are predicted in grains at any deformation path.

For this purpose different deformation histories were carried out under TEM micrograph as the following bellow:

1. Monotonic simple shear; simple shear 15% (stable orientation) and simple shear 30% (unstable orientation)
2. Reverse test; simple shear 30%/simple shear -30% (stable orientation) and simple shear 30%/simple shear -30% (unstable orientation)
3. Cross test; tensile 10% + shear 20%

For each TEM micrograph the grain orientation with respect to the macroscopic frame (RD, TD, ND) was known. Finally the local densities of immobile dislocations (pwd) stored in CBBs and directionally movable dislocations (pwp) associated with CBBs were plotted versus shear strain which were obtained from the Taylor model. The trend was also in good agreement with TEM micrograph data, see Figure 2-24 and Figure 2-25.

From the above TEM observation it can be concluded that CBBs are generated parallel to the active slip planes. This means that the number of

dislocation sheets families relates to the number of active slip systems. Also BCC crystal structure has only one family of unstable CBBs but in unstable one (discontinuous changes in strain path), two family of CBBs will generated. The other difference between stable and unstable orientation is that dislocation sheets are more polarized in the latter condition. Therefore the bauschinger effect and transient zone are more significant in an unstable orientation. From the Taylor prediction it was found that this model could be applied for qualitatively determining the intensity and the polarity of the dislocation sheets in a grain during any deformation path.

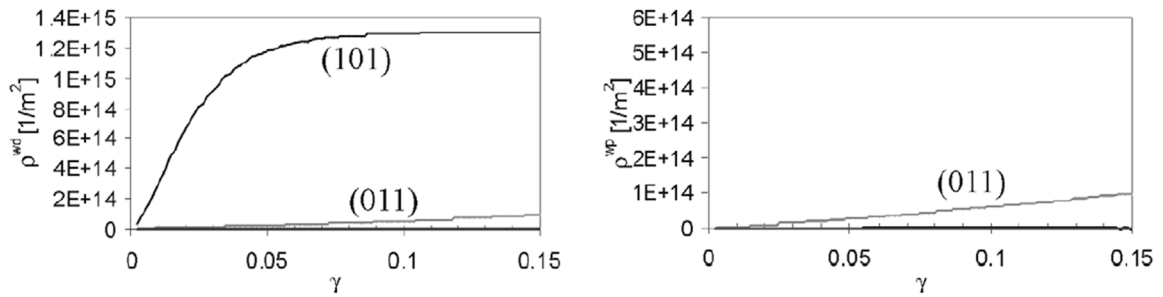


Figure 2-24: The evolution of the intensity and polarity of the CBBs in a crystal with initial orientation $(-27.2, 133.4, 53.0^\circ)$ during the simple shear test described in predicted by the extended Taylor model (68).

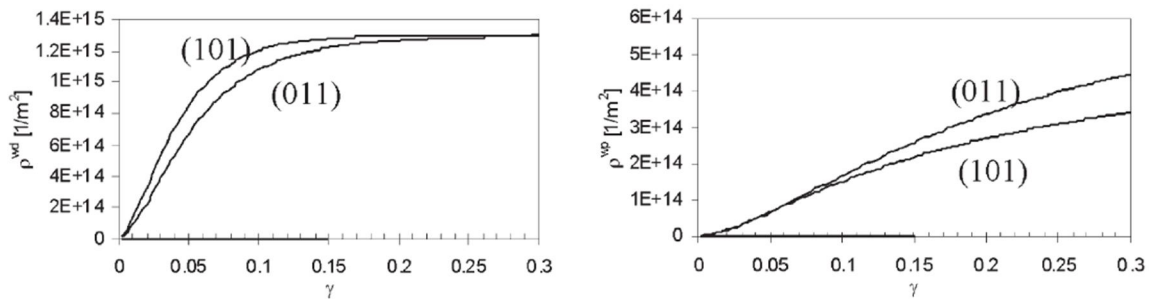


Figure 2-25: The evolution of the intensity and polarity of the CBBs in a crystal with initial orientation $(-44.5, 137.5, 42.0^\circ)$ during the simple shear test described in predicted by the extended Taylor model (68).

2.4.3 The effect of strain path changes on plastic slip anisotropy (Using cell structure model)

The material anisotropy development during pre-strain history is responsible for the macroscopic effect of a strain path changes. The following sources of anisotropy are associated with strain path changes during reloading: geometrical anisotropy (appearance of a dislocation structure), internal stresses, crystallographic slip and textural anisotropy (69).

The cell structure model for predicting the strain path changes effect was introduced by E. M. Viatkina et al. (70) but it didn't consider all sources of anisotropy. This problem encouraged E. M. Viatkina et al. (69) to work on the role of plastic slip anisotropy in the strain path changes modeling. Normally plastic anisotropy is related to strain path changes effect. Even in the highly symmetric FCC structure, all the slip systems do not cause an isotropic plastic deformation. Changing the reloading mode can affect the current microstructure. For this purpose, the cell structure model was adopted and consists of four uniform components: the cell interiors (low dislocation density area) and three sets of cell walls (high dislocation density area). The morphology of this composite material, used in this study is represented in Figure 2-26.

However employing crystal plasticity in the cell structure model increases computation time significantly, especially when polycrystalline simulation are carried. Yet it has a better prediction for the effect of strain path changes. The development of the internal stress components β_{12} in the cell structure during

monotonic shear simulation for grains with different lattice orientations is shown in Figure 2-27.

According to slip anisotropy, the internal stress inside the grains will be produced. Consequently it may have an influence on the strain path dependency in textured polycrystals. Thus for accurate prediction, slip anisotropy must be considered with dislocation structure and internal stresses.

Recently, several studies were performed on the development of cell structure models. A new approach for modeling of the elastoplastic behaviour of metallic materials under strain path changes was performed by Jamal Fajoui et al. (71). In this approach the grains were considered as a two-phase material with cell walls and cell interiors. First a micro-meso transition based on a modified elastoplastic self-consistent model which studies the dislocation evolution to obtain grain descriptions. Then a meso-macro transition using a self-consistent approach is utilised to deduce the macroscopic behaviour of the polycrystal. A good agreement between simulation predictions and experimental results was observed at the mesoscopic and macroscopic levels.

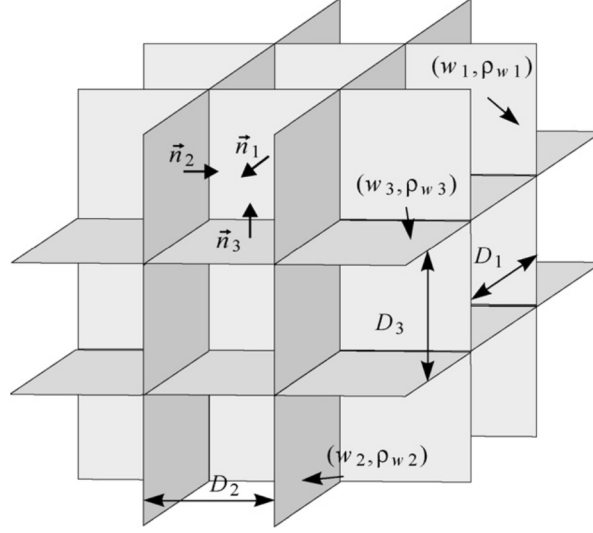


Figure 2-26: Model geometry of the cell structure and its components (69)

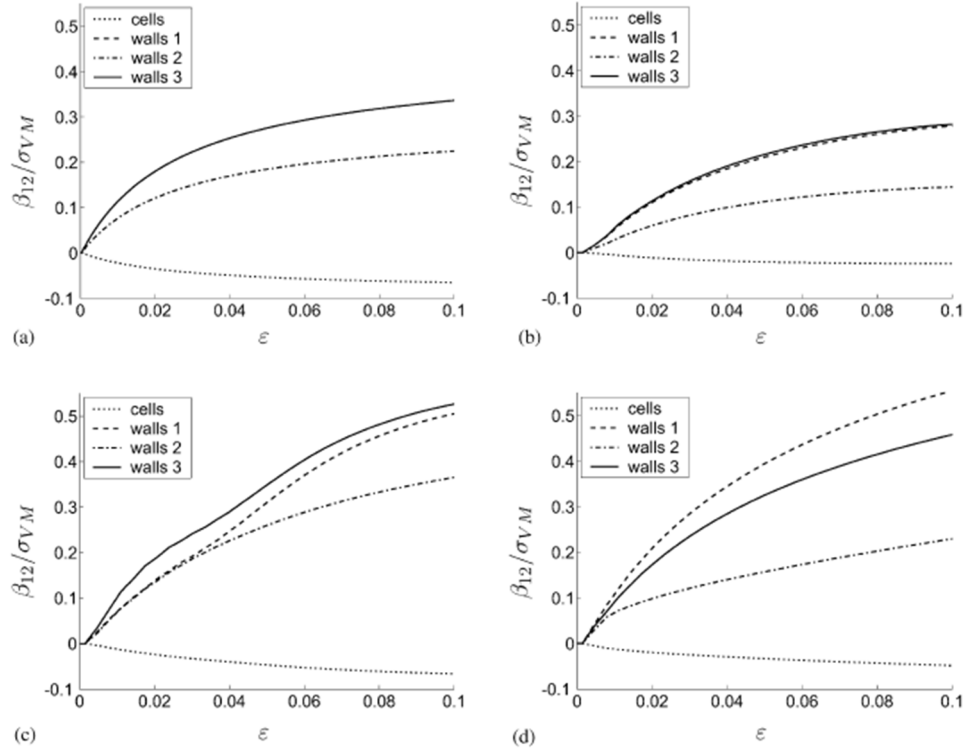


Figure 2-27: Evolution of the internal stresses β_{12} during monotonic shearing in the reference direction ($\xi=0^\circ$). (a) Predicted with the “continuum” model in Viatkina et al. (2007); Predicted with the present model for different grains, the grain orientations are given by the Euler angles. (b) Grain orientation $\{44^\circ, 113^\circ, 156^\circ\}$. (c) Grain orientation $\{67^\circ, 93^\circ, 95^\circ\}$. (d) Grain orientation $\{0^\circ, -135^\circ, -98^\circ\}$ (69).

T. Yalcinkaya et al. (65) worked on formation and evolution of these dislocations cell structure during strain path changes. They used a composite dislocation cell model which was combined with a BCC crystal plasticity framework to connect the microstructure evolution and its macroscopic anisotropic effects, see Figure 2-28. These models typically have many adjustable parameters that are generally fitted to obtain a good agreement between the model and the macroscopic behaviour. More quantitative experimental results at smaller length scales are needed to extract model parameters and to capture better predictions of strain path change effects.

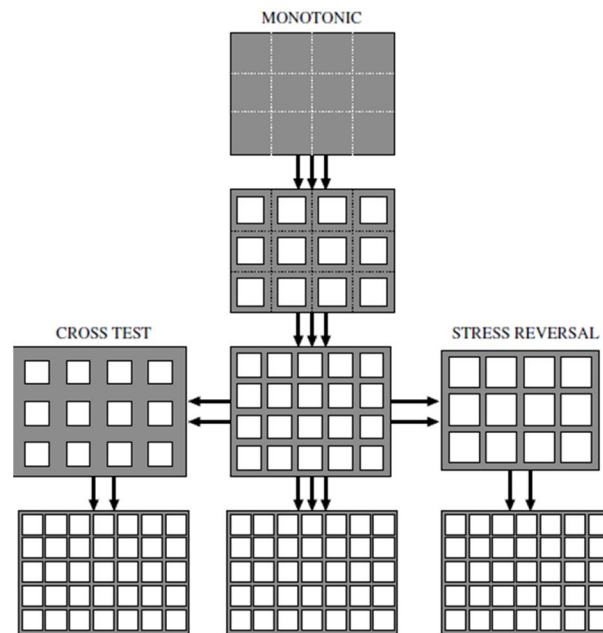


Figure 2-28: Schematic evolution of a dislocation cell structure under strain path change (65)

2.4.4 Effect of strain path changes on work-hardening/softening

2.4.4.1 Stress-strain Curve

Experimental studies of plastic deformation in BCC crystals show that dislocation distribution is not uniform during changing strain paths, and there are three main dislocation densities; layered structure, cellular structure and fragmented structure (72). Modeling of this non-uniform dislocation distribution and its effect on the stress-strain curve behaviour during plastic deformation is important in order to better understand sources of anisotropy and also it has important industrial applications.

A microstructural Taylor based model was developed in 2001 (72). This model took into account several parameters such as the formation of cell-block boundaries (CBBs), and the polarity of CBBs which are the main reasons for hardening and softening behaviour in stress-strain curves, see Figure 2-29.

The latent hardening mechanisms result in the formation of CBBs parallel to the most active slip systems, which implies that the rotation of the crystal or strain path changes are responsible for the activation of new slip systems and increasing the yield stress during plastic deformation. Therefore this jump in flow stress comes from the dislocation sheets and is named as “cross effect” (Δc in Figure 2-29).

The other effect of dislocations is the “Bauschinger effect” which causes a drop in flow stress (Δb in Figure 2-29). During plastic deformation, dislocations possess a mechanical charge (sign of burger vectors) causing a polarization of the CBBs. This polarization will make the slip resistance asymmetrical producing this drop in yield stress.

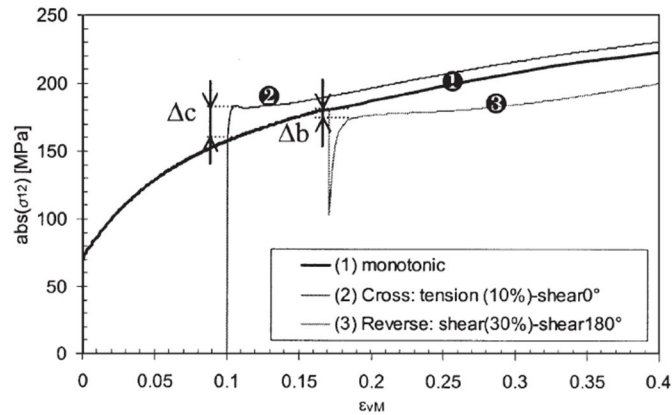


Figure 2-29: Experimental stress-strain behaviour of an IF steel during a monotonic simple shear test, a cross test and a reverse test (72)

This new extended Taylor model utilised kinetic equation for considering these defect densities in terms of hardening (or production) and softening (or recovery) and it was obtained by the separation of the net dislocation storage rate. However it was in good agreement with experimental data for IF steel during different two-stage strain paths; “tensile test/simple shear test”, “simple shear test/simple shear test”.

In addition to previous work there is a lot of research available on the effect of strain path changes on mechanical properties specifically for reverse strain. A drop in yield point in the stress-strain curve is expected according to the bauschinger effect, which is an elasto-plastic behaviour related to strain incompatibilities between grains that induce internal stresses (73).

In single-phase metals the stress-strain curve shows a rounded yield point on a strain reversal test. This behaviour is controlled by a low hardening rate until the

first inflexion point. After that the work hardening rate increases again until the second inflexion point, then it will decrease until rupture, see Figure 2.30.

This behaviour of single-phase metals has been seen in BCC steel and also some FCC metals such as Cu and Al (73).

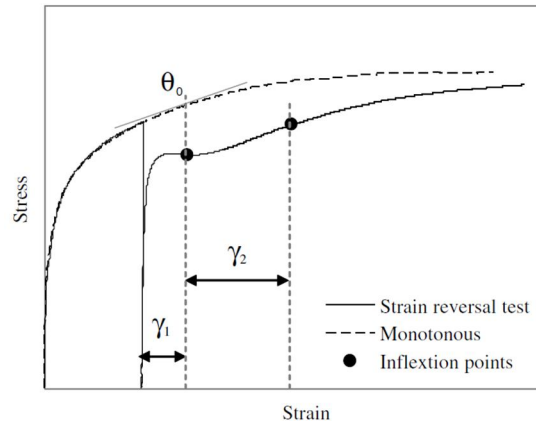


Figure 2-30: schematic stress–strain behaviour for single phase metal, subjected to a strain reversal test at large level of deformation (73)

Generally mechanical properties come from tensile tests at room temperature, but in industrial design or in some cold weather application it is necessary to know the appropriate plastic behaviour of the metals. G. Vincze et al. (73) carried out coupled reverse reloading and temperature changes tests on an AA1050 aluminum alloy and low carbon steel. For any temperature sequence the transient stagnation in the hardening rate was observed, stemming from dislocation substructure evolution. For example Figure 2-31 shows the shear stress-strain curves for AA1050 at room temperature and -120 °C under monotonic and reverse simple shear test. Also this larger stress in lower temperature is mainly related to the higher strain hardening rate.

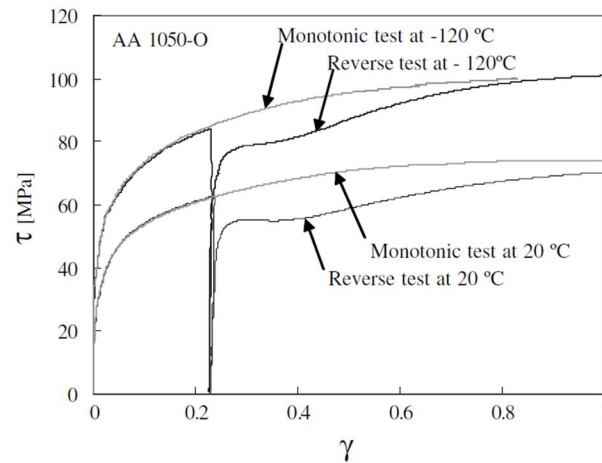


Figure 2-31: Forward and reverse simple shear tests performed on the aluminium alloy either at room temperature (lower curves) or at $-120\text{ }^{\circ}\text{C}$ (upper curves) (73).

The hardening stagnation effect and cross hardening effects mentioned above have been studied by S.H.A Boers et al (74) too. From the Figure 2-32, it can be seen that the evolution of dislocation microstructure has a strong effect on the material behaviour. For instance, as shown in case 3 in Figure 2-32, significant hardening followed by stagnation and then resumed hardening was observed under reverse shear loading after large amount of shear pre-strain. So, applying larger amounts of pre-strain caused an increase in hardening stagnation. In the orthogonal strain path the planar dislocation microstructure which forms in the first stage, acts as an additional barrier to the dislocation motion in the new loading direction. Then, after a significant jump in yield stress, the dislocations start to glide in the new direction leading to a breakdown of the old dislocation microstructure and decreasing the yield stress level back down to the monotonic level.

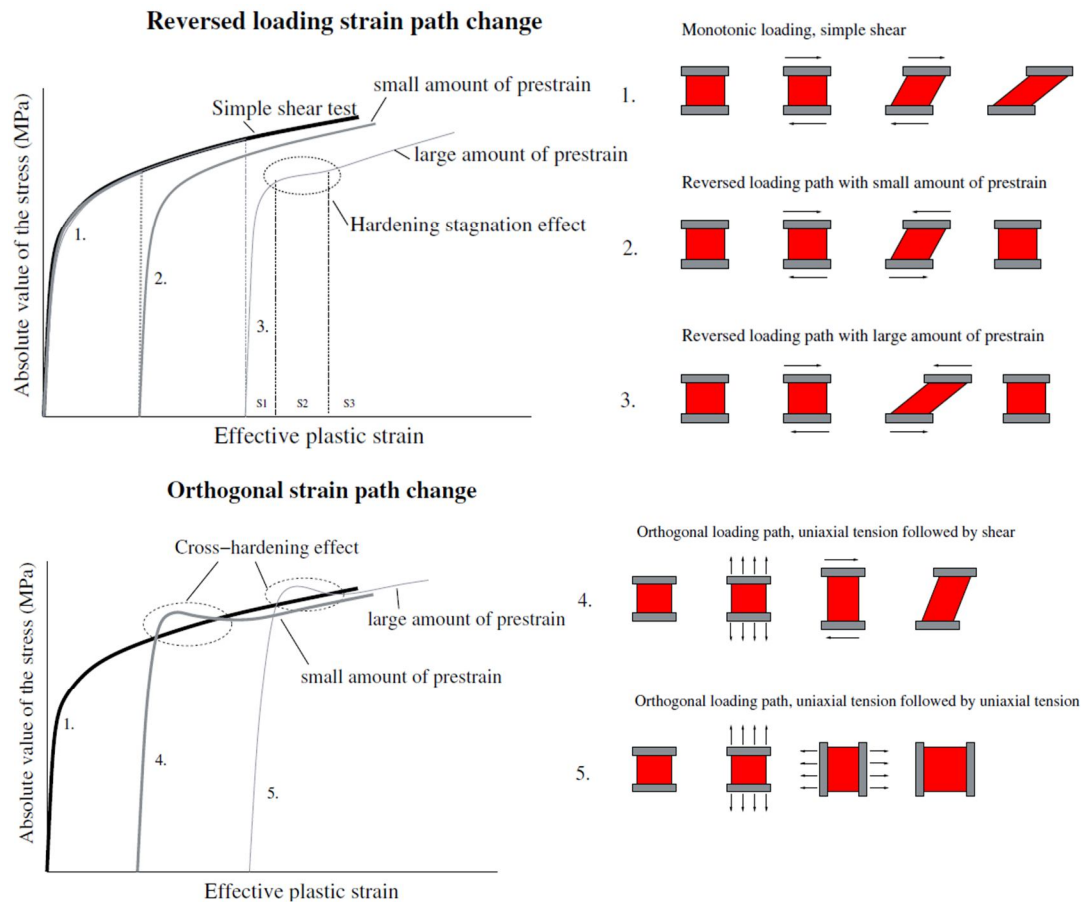


Figure 2-32: The effect of different loading paths on the macroscopic material response for monotonic, reversed and orthogonal strain paths (74)

2.4.4.2 Work-hardening rate

The above observation shows a strong effect of strain path changes on work hardening rate in metals. According to numerical approaches work-hardening rate is defined as $\theta = d\sigma/d\epsilon$, and is obtained from the slope of stress-strain curve. Work hardening rate or strain hardening rate is explained on the basis of dislocation-dislocation strain field interaction (75). The dislocation density increases by plastic deformation, so the average distance between dislocations decreases and they are located closed together. Consequently the resistance to dislocation motion by other

dislocations becomes more pronounced and more stress is required for further deformation. The evolution of the work-hardening rate depends on the nature of pre- and final deformation. It will be also be strongly affected by the change in strain path. Figure 2-33 shows the effect of strain path changes on work-hardening rate for highly pure copper (99.95%). This mechanical behaviour for two different types of path changes was studied by N.A. Sakharova, J.V. Fernandes (76).

For this purpose, the work hardening rate was plotted versus true stress as shown in Figure 2-33, for normal and parallel tension after rolling pre strain. A sharp drop of the strain hardening rate can be seen in both cases, after changes in strain path. Also after these drops, work hardening rates stabilize at a value close to 300MPa. But in the case of normal reloading an increase in work hardening rate after that sharp drop is significant which means that reloading tends to be recovered in tension. The result was in agreement TEM observations that showed the reorganization of the dislocation structure (76).

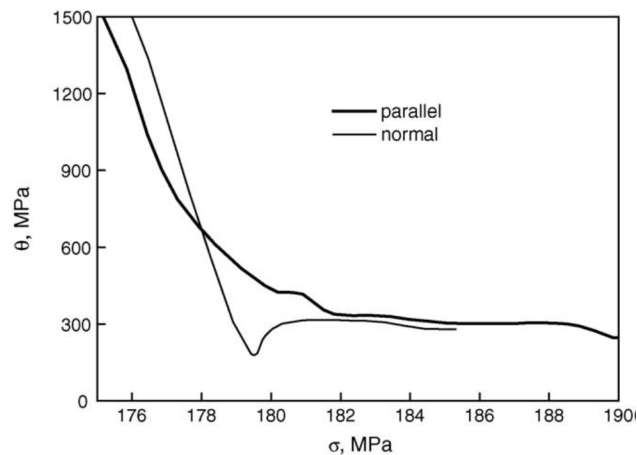


Figure 2-33: Work-hardening rate $\theta = d\sigma/d\epsilon$ vs. tensile stress σ , after normal and parallel strain path changes, for the rolling-tension sequence (76).

This work hardening rate behavior was confirmed by another study on 70/30 Brass sheet, deformed by simple tension, simple rolling and complex strain path of normal and parallel tensions after pre-straining in the rolling direction (77). Figure 2-34 shows the work-hardening evolution for brass samples with and without pre-straining in rolling.

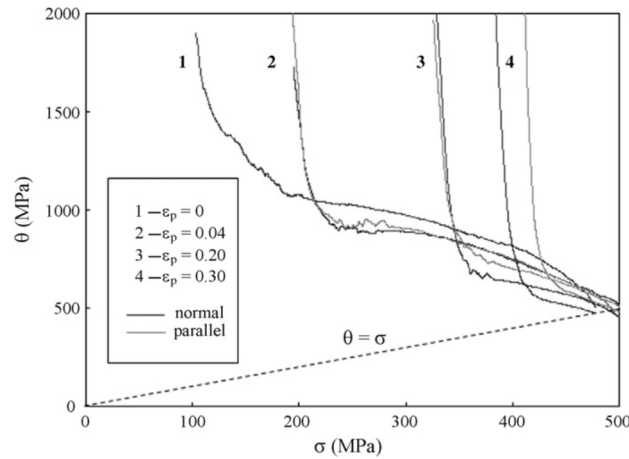


Figure 2-34: Work-hardening rate $\theta = d\sigma/d\varepsilon$ versus tensile stress σ for brass samples with and without prestrain in rolling direction (77).

From Figure 2-34 it can be seen that the work hardening rate has a relatively high initial drop (as for the material with high stacking fault energy; copper in the previous research), but this is followed by beginning of twin deformation (in material with low stacking fault energy; brass in current study). At some higher plastic deformations (more than 0.50) a second twin family has been observed.

2.4.5 The effect of strain path changes on forming limit diagram (FLDs)

A well-known method for explaining formability of metallic sheets is the Forming Limit Diagram (FLD) proposed by Keeler and Backofen (78). The FLD is generated using various monotonic proportional strain paths generally ranging

from biaxial tension (stretch forming) to equal tension and compression (deep drawing) and shows the major strain (ε_1) at the onset of localized necking for all values of the minor strain (ε_2).

An anisotropic constitutive model was applied to predict the forming limit diagrams (FLDs) by Shunji Hiwatashi et al. (79). This model is based on texture and dislocation structure. The FLDs prediction was studied for both proportional and non-proportional strain path. The value of limit strains is affected by mechanical properties of material such as initial anisotropy, transient hardening, cross hardening and Bauschinger effect. It has been shown that for proportional straining, limit strains near equi-biaxial stretching are significantly influenced by the initial texture, back stress and slip systems, Figure 2-35. Also the effect of strain path changes was studied for the following two-stage strain paths:

1. $\rho = -0.5$ to $\rho = 1$, i.e. axisymmetric tension followed by equi-biaxial stretching;
2. $\rho = -0.5$ to $\rho = 0$, i.e. axisymmetric tension followed by plane strain stretching;
3. $\rho = 1$ to $\rho = -0.5$, i.e. equi-biaxial stretching followed by axisymmetric tension;
4. $\rho = 1$ to $\rho = 0$, i.e. equi-biaxial stretching followed by plane strain stretching.

The authors could reproduce the experimental tendencies with the developed model, but they cannot be predicted by the conventional model. This is related to the occurrence of the transient hardening Figure 2-36.

Furthermore, the effect of first strain magnitude and the combination of the first and second strain modes on the limit strains was discussed (79). It was suggested that transient hardening significantly reduces the forming limit strain for a strain-path change from equi-biaxial stretching to uniaxial tension. Finally it was found that for more accuracy other factors such as texture evolution, damage, and strain rate sensitivity should be taken into account.

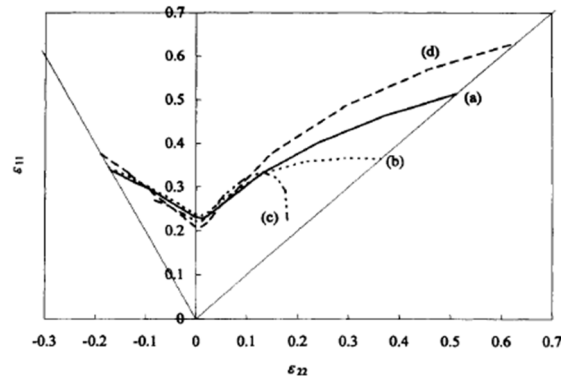


Figure 2-35: Forming limit curves for proportional straining predicted by (a) an anisotropic hardening model with a yield locus of mild steel where BCC type slip systems are assumed, (b) an anisotropic hardening model with a yield locus of a random-texture material where BCC type slip systems are assumed, (c) an anisotropic hardening model with a yield locus generated from the mild-steel texture where FCC type slip systems are assumed, and (d) an isotropic hardening model with the same initial yield locus as (a) (79).

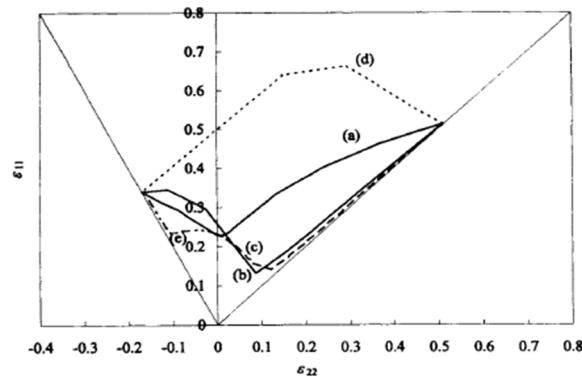


Figure 2-36: FLD for several strain paths, It is predicted with an anisotropic hardening model. Forming limit curves correspond to (a) proportional straining, and two-stage strain paths (b) from $p = 1$ to -0.5 , (c) from $p = 1$ to 0 , (d) from $p = -0.5$ to 1 , and (e) from $p = -0.5$ to 0 . The curves for non-proportional straining are generated by changing the amounts of the first strain (79).

2.5 Summary

The fundamentals of ductile fracture have been described with respect to both experimental and modeling views. Additionally the effect of pre-strain and strain path changes on different material properties has been reviewed. The ductile fracture processes (void nucleation, growth and coalescence) have been understood quite well from the modeling point of view, but the lack of experimental data has detrimental effect on validating the parameters affecting ductile fracture.

Metallic components may develop plastic deformation before in-service loading (pre-strain) due to manufacturing process and/or unexpected loading. This pre-strain not only affects the yield strength of the material but also influences its fracture properties. From the Table 2-5, it can be seen that tensile, mechanical behaviour and ductile fracture initiation were studied both experimental and using models.

It was shown that the development of plastic deformation is not always in the same direction and that metal can undergo complex loading sequences. These complex strain paths lead to reorganization of the dislocation structure, which generally results in the Bauschinger and cross effects which ultimately affect the stress strain curve and fracture properties.

2.6 Research objective

According to experimental evidences and modeling predictions in this literature review, more experimental data on ductile fracture mechanisms especially on void growth and void coalescence is required.

Although various models have been suggested to predict the effect of pre strain and strain path changes on material properties, these models generally have many adjustable parameters that are generally fitted to obtain a good agreement between the model and the macroscopic behavior. Therefore more quantitative experimental data at smaller length-scales are required to extract model parameters and to obtain better predictions of pre strain and strain path change effects. For this purpose the effect of pre-strain and strain path changes on hole growth and linkage was followed in-situ and compared with analytical and finite element models.

Chapter 3

Experimental Procedure

3.1 Introduction – Fabrication of model material

The purpose of this work is to better understand the ductile fracture process, particularly the growth and linkage between voids after different pre-strain and strain path changes histories. In order to better understand how the fracture process is affected by pre-strain and strain path changes, model materials containing fine laser drilled holes were utilized because the commercially available materials have a complex microstructure (random size, shape and distribution of inhomogeneity) which makes it difficult to isolate the parameter affecting fracture. Moreover, having the same initial configuration of voids (size, spacing and orientation) will allow better comparison between different pre-strain and strain path change histories.

3.2 Material selection

According to previous studies (39) (80) (81), AA5052 (purity: 96%-96.6%) has a high failure strain (~21 %) and work hardening rate along with a high ultimate tensile strength (~350 MPa). This material was received in the form of 100

μm thick sheets and contains intermetallic particles (up to 10μm in size) as shown in Figure 3-1. The micrograph shows the grain boundaries and intermetallic particles which were obtained after the etching process shown in Table 3-1. The chemical composition of aluminum alloy 5052 is shown in Table 3-2. The 5XXX series of aluminum (5xxx) are magnesium-based alloys. Due to their formability, corrosion resistance and weldability these alloys are commonly applied in pressure vessels, fuel and oil tubing, trucks bodies and electrical cabinets.

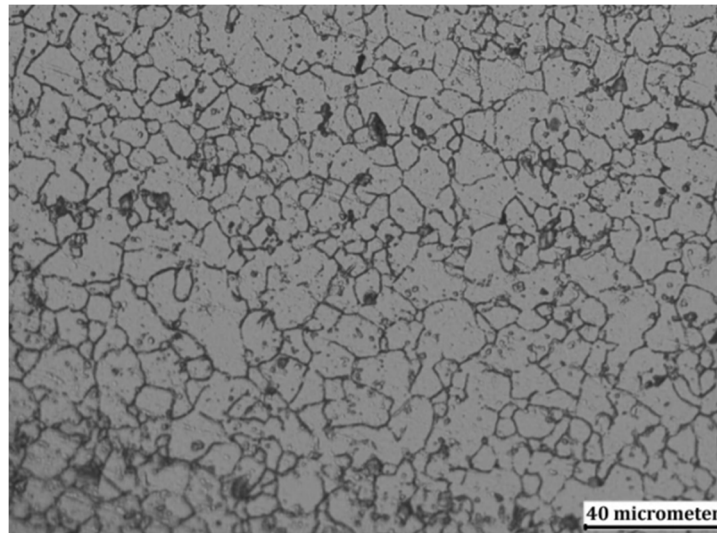


Figure 3-1: Optical micrographs of aluminum alloy 5052 showing the intermetallic particles and grain size (after etching)

Table 3-1: Etching procedure of 5052 aluminum alloy samples

Sample	Heat treatment	Solution	Time	temperature
5052 aluminum alloy	Heat treated at 125°C for 3 days to segregate the Mg at the interfaces	50% H ₃ PO ₄ + Warm water	2 minutes	45 °C

Table 3-2: Chemical composition of the aluminum magnesium alloy 5052

Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Other	Al
0.25	0.40	0.10	0.10	2.2-2.8	0.15	0.10	0.05	0.015	Rest

3.3 Pure pre-strain tests (pre-strain and strain to failure in the same direction)

3.3.1 Pre-strain samples geometry

In order to study the effect of pre-strain history on ductile fracture, small dog bone shaped samples were extracted from a 100 μm thick sheet of 5052 aluminum alloy, see Figure 3-2. Sample dimensions were chosen to fit in the small in-situ tensile machine used to carry out the tensile tests under the optical microscope.

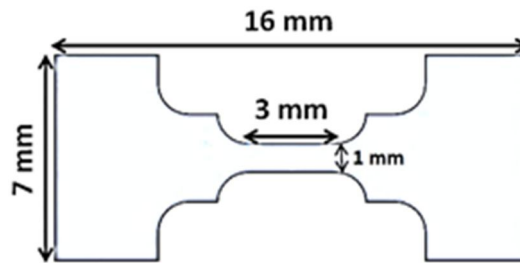


Figure 3-2: Schematic illustration of a small tensile sample (all the dimensions are in mm)

3.3.2 Heat treatment (annealing)

The as-received material may contain some prior deformation due to machining so prior to pre straining the specimens are heat-treated at 345°C for 30 minutes to remove the hardening effect of accumulated dislocations. For this purpose the low vacuum furnace (pressure = 5×10^{-3} to 10^{-4} torr) was used as is shown in Figure 3-3.

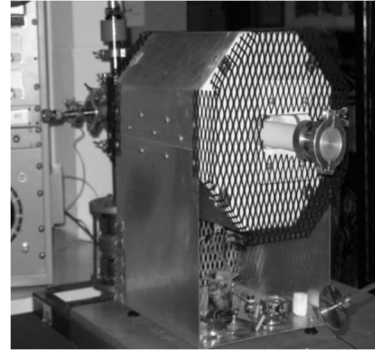
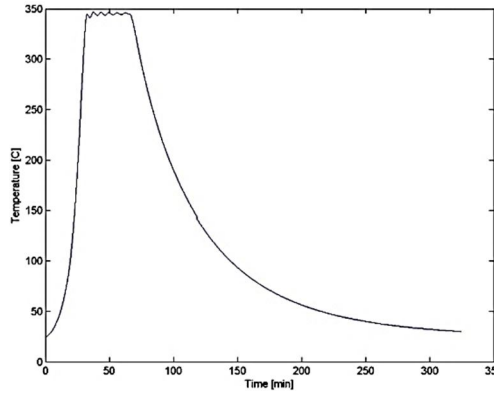


Figure 3-3: Heat treatment of 5052 aluminum alloy at 345 °C for 30 minutes; a) annealing Temperature versus time, b) low vacuum furnace (pressure = 5×10^{-3} to 10^{-4} torr)

3.3.3 Pre-strain tests

In order to impose pre-strains, the tensile samples were subjected to tensile deformation at a constant crosshead speed of $0.893 \mu\text{m/s}$. Deformation was continued until a pre-determined total strain level was achieved, after which the samples were unloaded. Pre strains of 4.2%, 8.1%, 12.75%, 17.25% and 21.45% were applied through this process. Figure 3-4 shows the tensile machine with 100lb load cell where its strain rate was calculated from Equation 3-1 as follows:

$$\text{Strain rate} = \frac{(\text{speed of crosshead})}{(\text{gauge length of small tensile sample})} = \frac{0.893 \frac{\mu\text{m}}{\text{s}}}{3000 \mu\text{m}} = 3 \times 10^{-4} \quad \text{Equation 3-1}$$

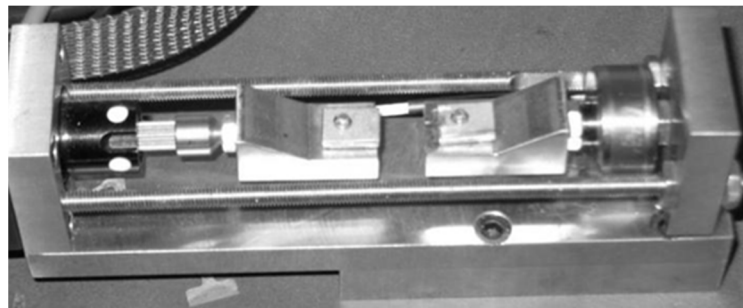


Figure 3-4: 100lb load cell tensile machine (Zaber NA14B16-V); using for uniaxial tension test with strain rate of 3×10^{-4}

3.3.4 Effect of tensile machine rigidity

A common misconception is that the tensile machine is more rigid than the tensile specimen. In fact the opposite is often true and the tensile machine crosshead displacement is then the sum of the deformation in the gauge length of the specimen plus the elastic deflections in components such as the machine frame, load cell, grips, and specimen ends (82). Determining the rigidity factor of the tensile machine is required in order to obtain the real displacement in the sample. For measuring the rigidity factor tensile tests were carried out in-situ under an optical microscope and pictures were taken every second during the tests. Results from the tensile machine were compared to the corresponding real values which were obtained by tracking some features on the sample surface.

Figure 3-5 represents the cross head displacement versus the real specimen displacement and the linear fit provides the stiffness factor of the tensile machine. The value of 1.5 from Figure 3-5 was taken into account for calculating the real pre-strain values. So the real pre-strain values are 2.8%, 5.4%, 8.5%, 11.5% and 14.3%.

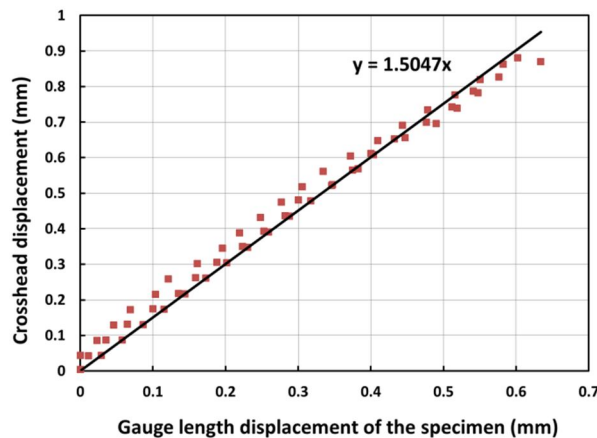


Figure 3-5: Measuring the rigidity factor by comparing the crosshead displacement versus gauge length displacement

3.3.5 Laser hole drilling of the samples

3.3.5.1 Laser micromachining set-up

The laser used to create the holes in the tensile coupons was a Newport Spitfire Pro amplifier coupled with a Newport Tsunami femtosecond laser oscillator. The main advantage of femtosecond laser is that they create a negligible heat affected zone around the outside of the void. Additionally they can easily create micron-sized features (83). The laser system had a wavelength of 800 nm, a frequency of 1 kHz and a pulse width of less than 40 fs. The laser was coupled to a machining set-up with an x-y computer controlled stage coupled to a Matlab program for precise control of the machining process (see Figure 3-6).

Holes were created in the sample gage lengths by focusing laser pulses at the surface of the sample through a 10x objective lens. The laser pulses had a total energy of 23 μJ and an exposure time of 3 sec per hole.

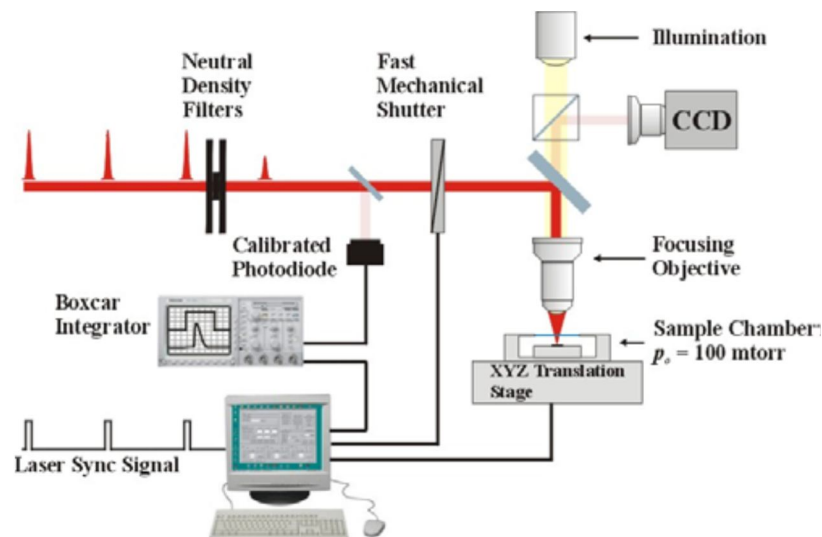


Figure 3-6: Laser micromachining setup (1)

3.3.5.2 Hole diameter selection and hole configuration

As mentioned in section 3.1, one of the main parameter that has a significant effect on driving the fracture processes is void geometry. So the size of the laser-machined holes, their spacing and their distribution are very important parameters for studying void growth and void coalescence processes. Moreover, hole sizes in the range of a few microns to tens of microns should be selected for reproducing the microstructural features found during fracture of commercial aluminum alloys. However the void dimensions must also be applicable with the characterizing technique that is used in 2D tensile experiments under an optical microscope. Finally a void diameter of approximately 20 μm was selected for the current study. Note that the actual diameter was 17.5 μm due to difficulties in obtaining exactly 20 μm during laser machining and because of subsequent polishing steps.

As the primary goal of this project is to study the effect of pre-strain and strain path changes on both void growth and void coalescence, more than one hole is needed to see the linkage between the holes. Thus in the center of the sample an array of 10 holes with 35 μm interspacing was laser drilled through the tensile sample for better understanding the linkage between the holes. Having at least 0.2 mm of hole-free material on each side of the array helps better control the final fracture. Additionally the 35 μm holes interspacing (twice the hole diameter) allows for enough void growth prior to failure. The optical micrographs showing the holes configuration of the samples with 10 holes is shown in Figure 3-7.

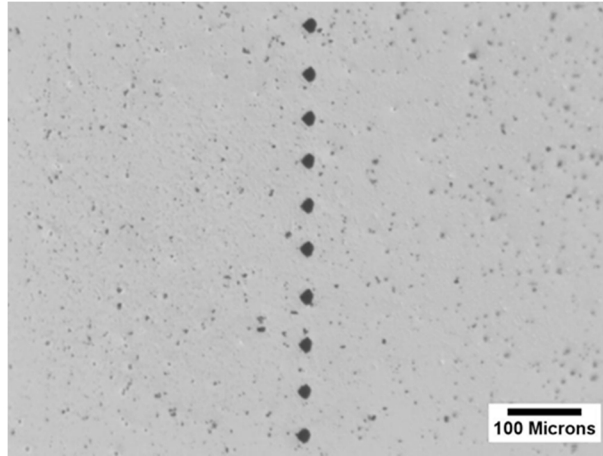


Figure 3-7: Optical micrograph of polished 5052 aluminum alloy showing 10 holes in the center of the sample

3.3.6 Sample preparation for optical observation

Prior to carrying out the final tensile tests to failure, samples have to be polished for better observations in the optical microscope. The polishing technique creates a smooth and shiny sample by grinding away surface material repeatedly with initially coarse sandpaper and proceeding to finer cloths. The difference in surface after polishing is shown in Figure 3.8. From this figure, it can be seen that the surface features from laser drilling and rolling marks were removed during surface polishing. Also note the decrease in void diameter after polishing. The polishing procedure is shown in Table 3-3.

Table 3-3 Polishing procedure of 5052 aluminum alloy

Cloth	Time
4000 SiC paper	2 minutes or until the surface is uniform
0.05 μm paper (very fine surface)	1min with colloidal silica suspension, add spurts water periodically, leave water running for last ten seconds

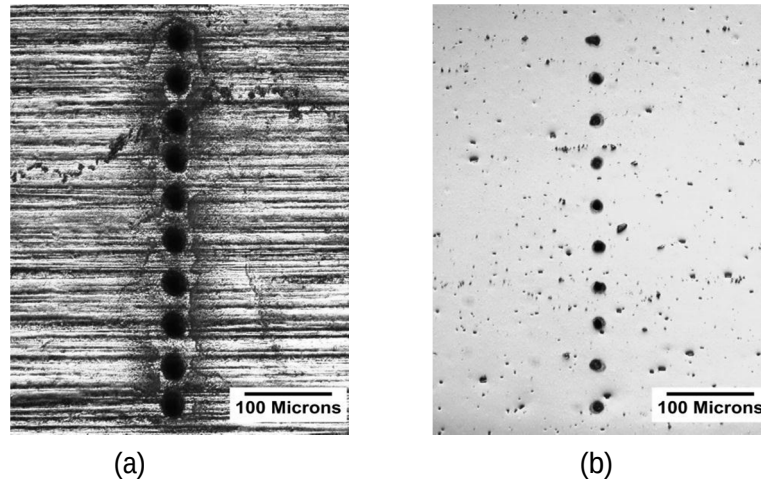


Figure 3-8: Optical images of 5052 aluminum alloy showing the difference in surface before and after polishing; a) before polishing b) after polishing

3.3.7 In-situ tensile testing to failure under optical microscope

Final tensile tests to failure were carried out at a cross-head speed of $0.893\mu\text{m/s}$ with the samples being pulled in-situ under an optical microscope which allowed pictures to be taken during the test every 3 seconds. Figure 3.9 shows a typical optical microscope image of the sample prior to deformation and right before failure. Both load and displacement were recorded during the tensile test and the displacement was corrected to account for the tensile machine rigidity. Then the engineering stress-strain curves were obtained by dividing the load-displacement by the cross sectional area and gauge length respectively. Since it is more meaningful to use true stress-true strain curves, as they will be applied as input in the finite element modeling (chapter 5), all engineering stress-strain curves are converted to true ones. Figure 3-10 shows a typical true stress-strain curve for aluminum alloy 5052 tensile sample including both pure pre-strain and strain path changes tests which are described in the section 3.4.

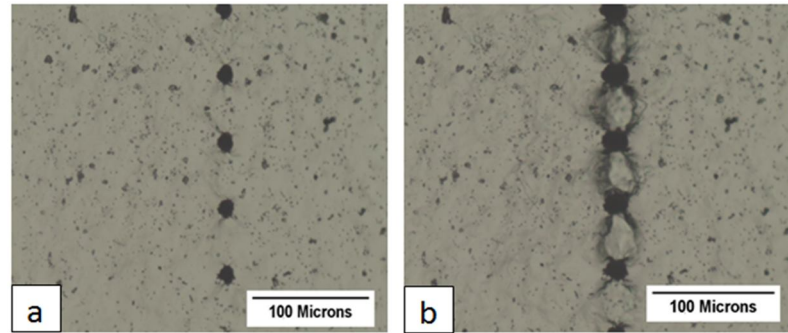


Figure 3-9: Optical images showing the void growth and coalescence in a line of holes oriented at 90° with respect to the horizontal tensile axis; a) beginning of the deformation b) final state, before fracture

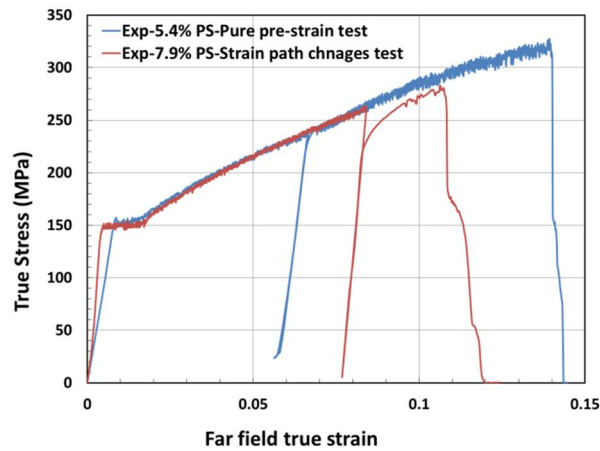


Figure 3-10: Plot of true stress versus far field true strain for aluminum alloy 5052 for pure pre-strain (blue line) and strain path changes (red line) tests

3.3.8 Image analysis

From the optical microscope images, the major diameter of the voids was extracted at various deformation steps. Then the evolution of the normalized void radius (a/a_0) where a_0 is the initial void radius, as a function of the true far field strain defined as $\ln(1 + e_{ff})$ where e_{ff} is the far field engineering strain was plotted for different pre-strain histories. Due to the large number of microscope images (>100) and for repeatability, void lengths were extracted using an open-source

image analysis software (ImageJ). Figure 3-11 shows the schematic drawing of the void dimension and intervoid spacing parameters for the tensile specimen.

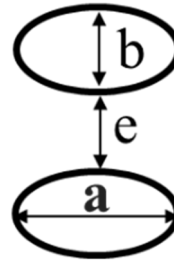


Figure 3-11: Schematic drawing of Void dimension parameters normal to the tensile axis; a=major radius, b=minor radius and e=distance between two voids

This image analysis software (Image J) contains options that adjust brightness/contrast, threshold levels and image size. By selecting the “threshold” tool, lower and upper threshold values, segmenting grayscale images into features of interest and background can be set automatically or interactively. Figure 3-12 briefly represents the Image analyzing procedure. The large amount of experimental data (more than 100 images per test) that can be processed and repeatability in the measurements are the main advantages of this technique compared to manually measuring hole dimensions.

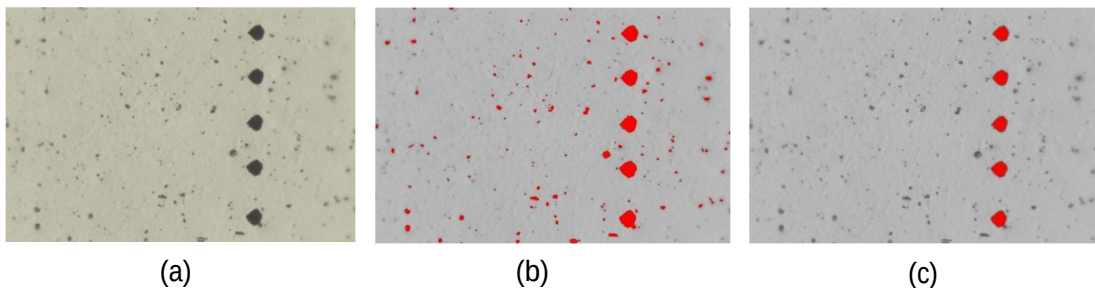


Figure 3-12: Image analysis software processes; a) Uploading the raw image, b) Thresholding the image, c) set the lower and upper threshold for final hole dimension measurement

Comparison between manual image analysis and results obtained with ImageJ is shown in Figure 3-13, Figure 3-14 and Figure 3-15. The experimental data extracted with Image J are in good agreement with manual measurements which validates the image analysis technique. All subsequent image analysis was performed using Image J.

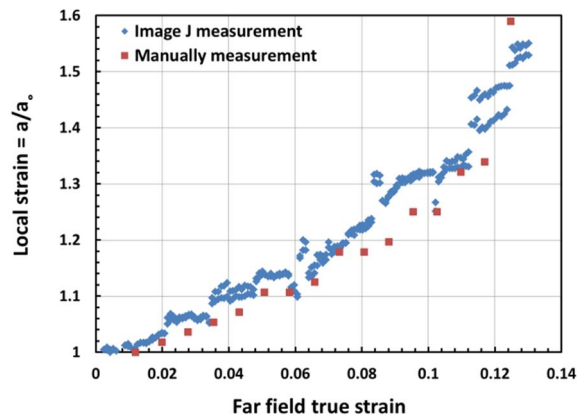


Figure 3-13: Local strain versus far field strain for the 5052 aluminum alloy with 10% pre-strain; comparison between Image J measurement and manually measurement

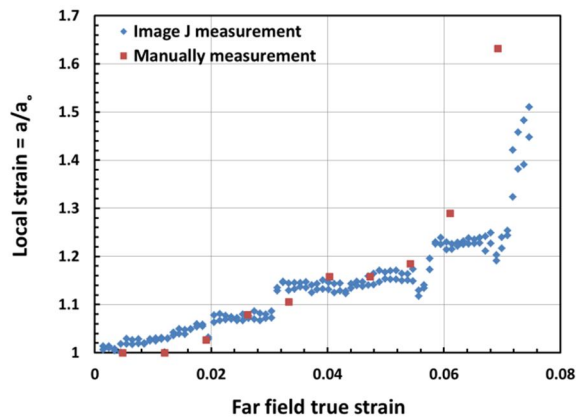


Figure 3-14: Local strain versus far field strain for the 5052 aluminum alloy with 20% pre-strain; comparison between Image J measurement and manually measurement

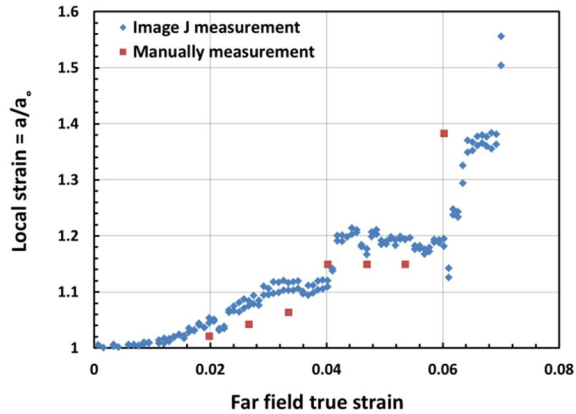


Figure 3-15: Local strain versus far field strain for the 5052 aluminum alloy with 25% pre-strain; comparison between Image J measurement and manually measurement

3.3.9 Fracture surface analysis in scanning electron microscope (SEM)

Finally fracture surface analysis in a Scanning Electron Microscope (SEM) was carried out to determine the true strains at fracture (ductility) for 5052 aluminum alloy tensile specimens. For this purpose the JEOL JSM-5300 Scanning Electron Microscope (SEM) was utilized at a working voltage of 20 keV (Figure 3-16).



Figure 3-16: JEOL JSM-5300 Scanning Electron Microscope (SEM); Resolution: 10-20 nm, Magnification: high vacuum mode 15x-200,000x and low vacuum mode 15x-50,000x, Acceleration voltage: 0.5-30 kV, Specimen size: up to 76 mm in diameter.

3.4 Strain path changes tests; experimental procedure

3.4.1 Strain path change samples

For obtaining the effect of strain path changes on ductile fracture, large tensile specimens were extracted from a 100 μ m thick 5052 aluminum alloy sheet as shown in Figure 3-17.

3.4.2 Heat treatment

The same annealing parameters as in section 3.3.2 were used for the large tensile samples. They were annealed at 345 °C for 30 minutes in the low vacuum furnace.

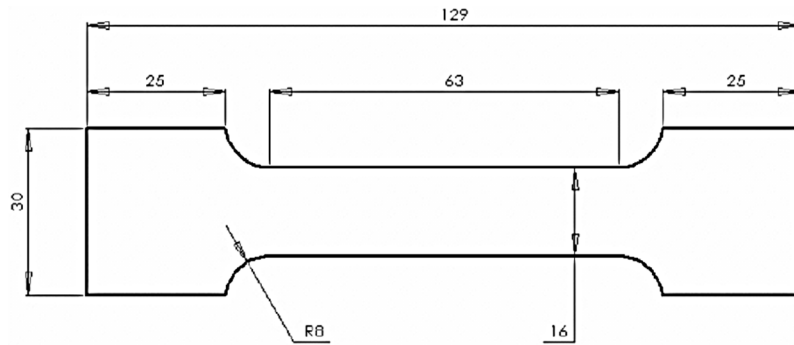


Figure 3-17: Schematic illustration of big tensile sample (all the dimensions are in mm)

3.4.3 Pre-strain test coupled with digital image correlation (DIC)

3.4.3.1 Digital image correlation (DIC)

Digital Image correlation is an optical method to obtain 2-D full field information by recording the motion of speckle patterns on a specimen surface during deformation of a body. DIC tracks the position of the same physical points shown in a reference image and a deformed image.

To achieve this, a square subset of pixels is identified on the speckle pattern around a point of interest on a reference image and their corresponding location determined on the deformed image. The boxes shown in Figure 3-18 are an example of the subset. The maximum correlation in each window corresponds to the displacement, and this gives the vector length and direction for each window.

The digital images are recorded and processed using an image correlation algorithm. Many parameters are included while obtaining accurate DIC results. Some of the parameters include speckle size, speckle density, type of algorithm, subset size, subset overlap, gray level interpolation. For the current study the Davis 8.1 software (complete software for intelligent imaging applications) was utilized.

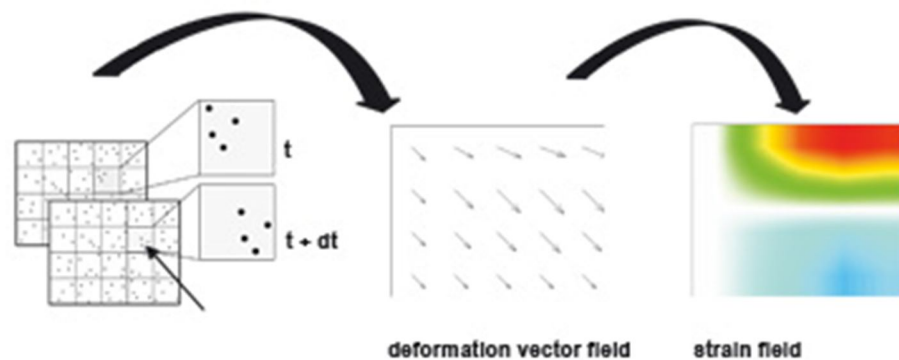


Figure 3-18: Digital Image correlation processes; strain evaluation (Image is taken from La Vision company website)

3.4.3.2 Tensile tests on large samples and DIC analysis

Prior to carrying out the pre-strain tests the large 5052 aluminum alloy tensile specimens were sprayed with black and white pattern as shown in Figure 3.19 (for DIC purposes).

Then the large tensile samples were subjected to tensile deformation at a constant crosshead speed of 1.13mm/min and the pictures were taken during the

test with a digital camera every 3 sec, see Figure 3-20. The same strain rate as pure pre-strain test was used as follows:

$$\text{Strain rate} = \frac{(\text{speed of crosshead})}{(\text{gauge length of small tensile sample})} = \frac{1.13 \frac{\mu\text{m}}{\text{s}}}{63 \mu\text{m}} = 3 * 10^{-4} \quad \text{Equation 3-2}$$

Deformation was continued until a pre-determined total strain level was achieved, after which the samples were unloaded. Pre strains of 5%, 10%, 15%, and 20% were applied through this process.

Since the pre-strain test was carried out in-situ, all digital images were uploaded in DaVis 8.1 software to obtain the real local strain value along the gauge length of the sample. Then small tensile samples were extracted from the gauge length of the large tensile samples using the femtosecond laser. The DIC results from the large tensile samples allow extraction of the exact pre-strain value in each coupon. The real pre-strain value which were obtained from DIC technique were 4.2%, 7.96%, 11.8% and 15.8%.

Figure 3-21 shows the strain field distribution for a large tensile sample and also the location for extraction of the small tensile samples.

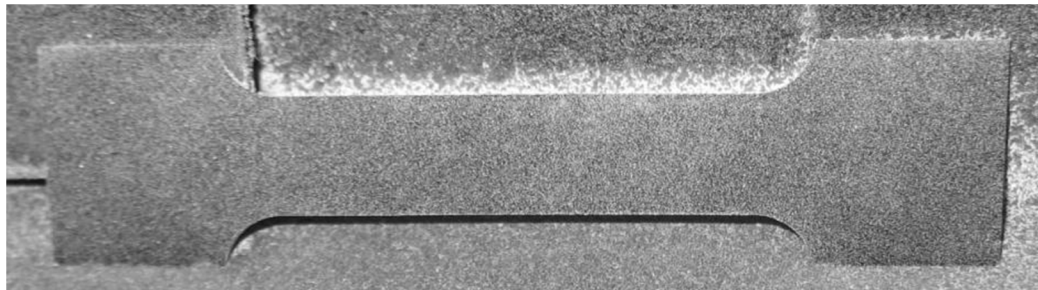


Figure 3-19: 5052 aluminum alloy tensile sample sprayed with black and white dot pattern



Figure 3-20: Pre-strain test setup; using tensile machine (Instron 4482) and taking picture during the test with digital camera

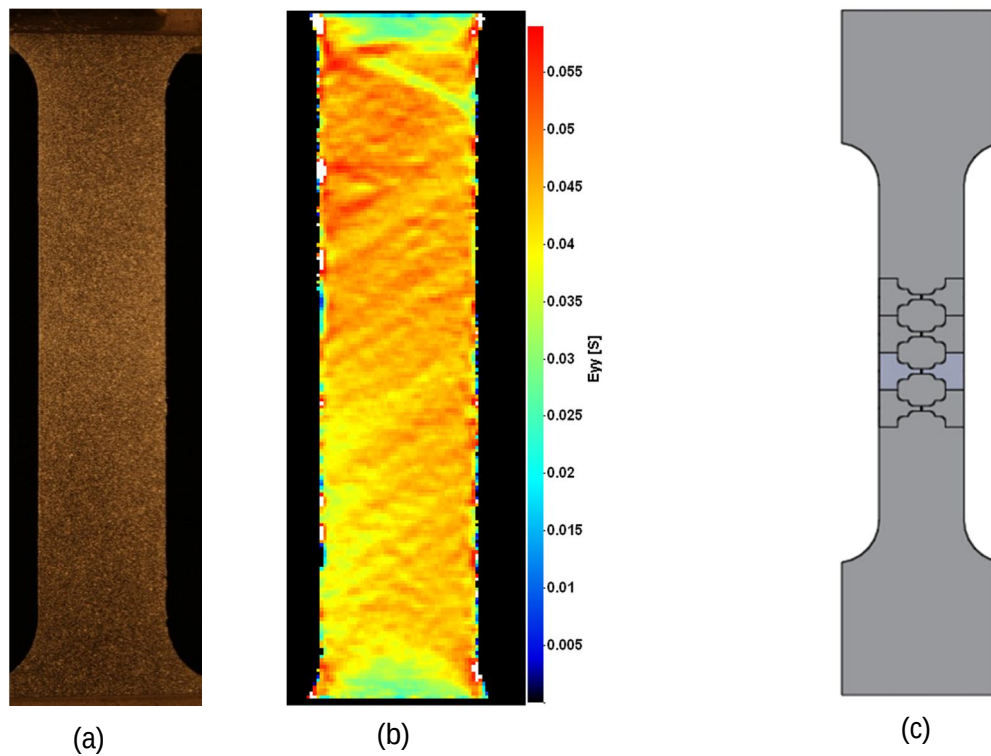


Figure 3-21: Digital image correlation technique for 5052 aluminum sample; a) raw image b) strain field distribution for big tensile sample after image processing and c) extracting small tensile samples procedure by laser machining

3.4.4 Orthogonal test to failure coupled with DIC

3.4.4.1 Electro deposition to create a speckle pattern for DIC

Electro deposition is a coating process in which metal ions in a solution are moved by an electric field to coat an electrode. The part to be coated is the cathode of the circuit. Both components are immersed in a solution called an electrolyte containing one or more dissolved metal salts as well as other ions that permit the flow of electricity. A power supply provides a direct current to the anode, oxidizing the metal atoms that comprise it and allowing them to dissolve in the solution. At the cathode, the dissolved metal ions in the electrolyte solution are reduced at the interface between the solution and the cathode, such that they "plate out" onto the cathode, see **Figure 3-22**.

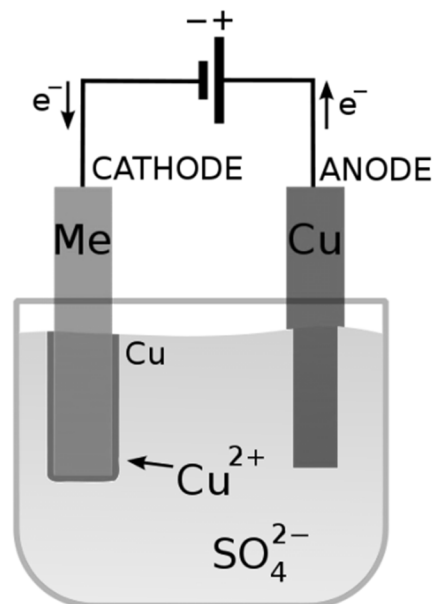


Figure 3-22: Schematic drawing of Electroplating of a metal (Me) with copper in a copper sulfate bath (Dr. Torsten Henning; Copper electroplating.svg)

The main advantage of this method compared to black and white spray pattern deposition is that the size can be controlled down to the micro scale. The size and density of the features on the metal surface are controlled by time and voltage respectively. For electro deposition of 5052 aluminum alloy a cupric sulfate solution was used for all experiments.

The electrolyte includes 40 g/L Cu (0.63 M) (Fisher Scientific in the form of $\text{CuSO}_4 \cdot 5\text{H}_2\text{O}$) and 10 g/L H_2SO_4 (0.10 M) (EMD Chemicals). Poly ethylene glycol (PEG) with 3350 g/mole molecular weight provided from Fisher Scientific was used in the experiments as the suppressor (84). By setting the voltage of the power supply to 40V, and a deposition time of less than 0.5 sec and repeating this procedure 20 times, a high density of fine features was obtained between holes with a feature size less than 2 microns , see Figure 3-23.

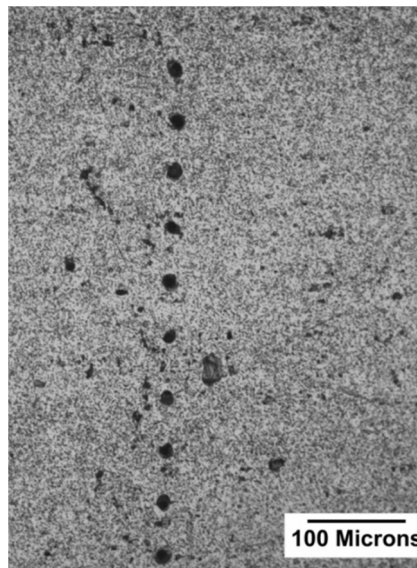


Figure 3-23: Optical image from 5052 aluminum alloy sample showing the high density of the features between the holes with less than 2 microns in diameter; the picture is taken after electro deposition in cupric sulfate bath (40 V – 0.5 sec; repeat more than 20 times)

3.4.4.2 Orthogonal test to failure

Final tensile tests to failure were carried out at a cross-head speed of $0.893\mu\text{m/s}$ with the samples being pulled at 90 degrees with respect to the pre-strain direction. The test was carried out in-situ under an optical microscope which allowed pictures to be taken during the test every 3 seconds.

The optical images obtained were then uploaded in DaVis 8.1 software for determining the local strain field between the holes. Figure 3-24 (d) shows the strain distribution between the holes for 5052 aluminum alloys. Since the strain path changes tests were carried out by another tensile machine, the new rigidity factor has to be determined and taken into account. The same procedure as pure pre-strain test was used and a stiffness factor of 1.35 from Figure 3-25 was found for new tensile machine. Figure 3-26 shows an image from new tensile machine.

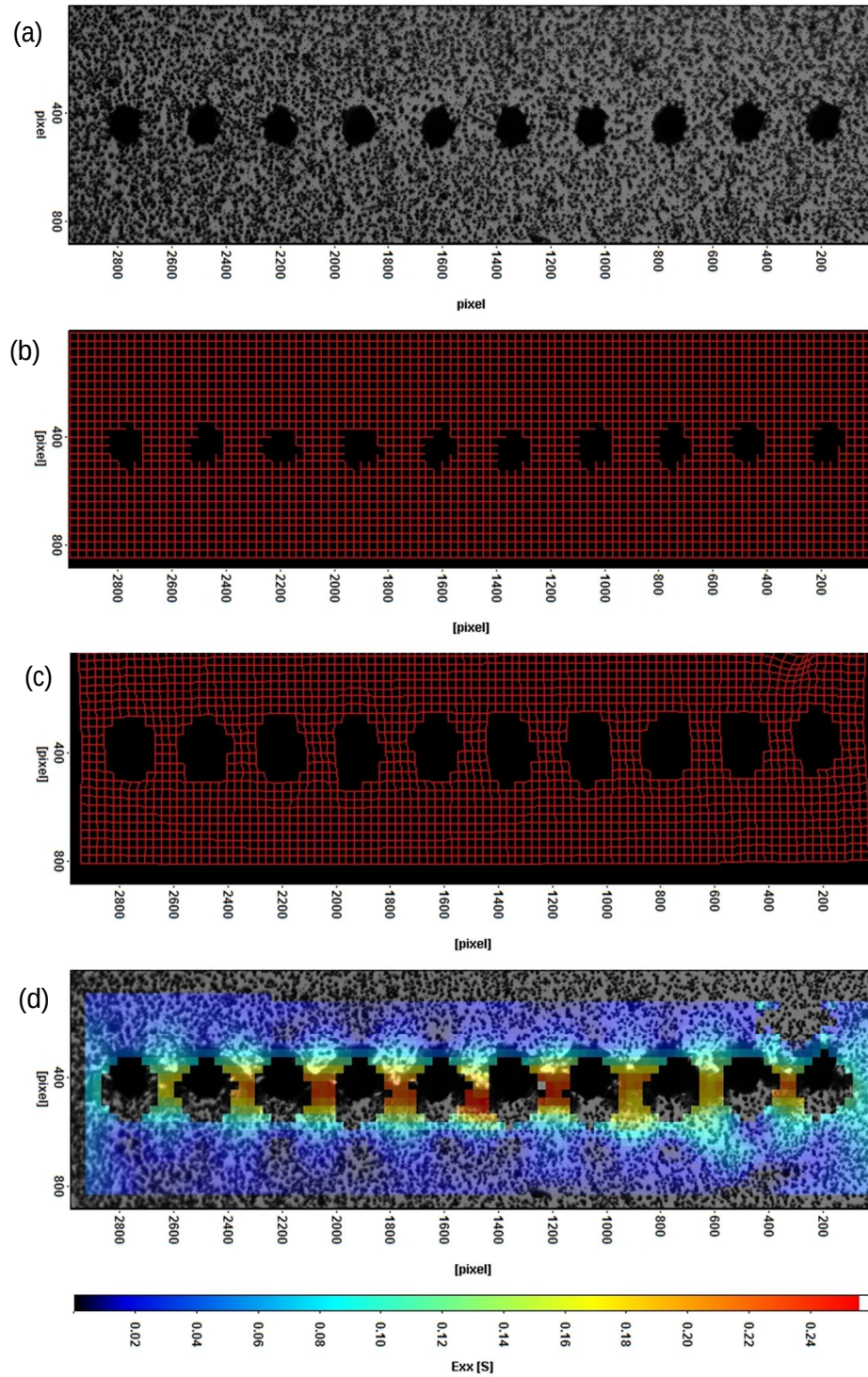


Figure 3-24: Digital Image processing steps for 5052 aluminum alloy tensile coupon; a) raw image b) grid field for the first image (before deformation) c) grid field for the last picture (after deformation) and d) 2D strain field distribution between the holes

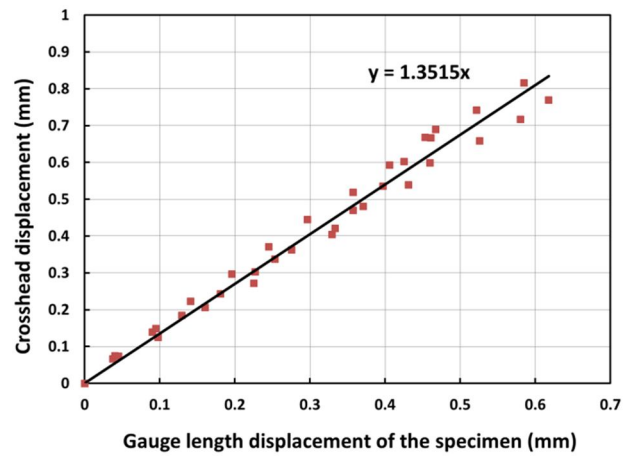


Figure 3-25: Measuring the rigidity factor by comparing the crosshead displacement versus gauge length displacement; new tensile machine

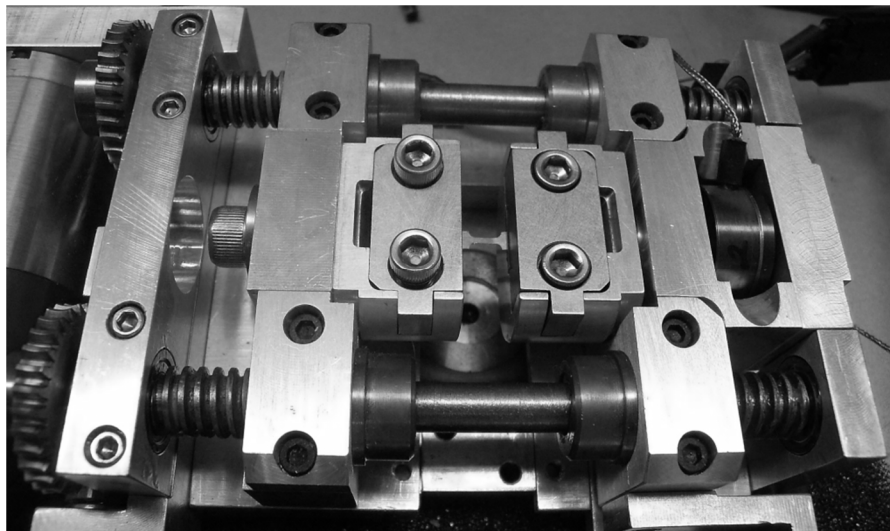


Figure 3-26: 100lb load cell tensile machine (FAULHUBER); using for uniaxial tension test with strain rate of 3×10^{-4}

Chapter 4

Experimental results and Discussion

4.1 Introduction

The main goal of this chapter is to show that it is possible to experimentally visualize, in a controlled and reproducible way, the effect of pre-strain and strain path changes on void growth and void coalescence through in-situ tensile tests on laser drilled aluminum alloy 5052. By controlling the position, size and quantity of voids, the effect of pre-strain and strain path changes on the growth and coalescence of voids can be studied. Moreover, with obtaining the major diameter of voids as a function of deformation, the accuracy of the McClintock model (presented in the literature review) and validation of finite element models (presented in chapter 5) can be assessed.

The first part of this chapter presents the effect of pre-strain on the stress-strain curve, ductility of the material and void growth and coalescence. At the end of this part the experimental results are compared with the McClintock model for void growth. The second part deals with the effect of strain path changes on stress-strain curves, on void growth and coalescence using digital image correlation map. In the last section pre-strain and strain path changes results are compared to each other.

4.2 The effect of pre-strain history on ductile fracture

As it was shown in chapter 2 most of previous works on pure pre-strain (without strain path effect) discussed the effect of pre-strain on either microstructure (such as grain boundaries and dislocations) or macrostructure (such as fracture toughness and hardening exponent). Modeling approaches for void growth and coalescence have been proposed but the lack of experimental data in this area results in the poor prediction capability of these models. In this section the effect of pre-strain on mechanical properties of aluminum alloy 5052 is shown first. Then changes in ductility upon pre-straining are presented from an analysis of SEM images of the sample's fracture surface. Finally the effect of pre-strain on void growth and coalescence is demonstrated by measuring the major void diameter during in-situ tensile tests, along with a comparison of experimental results with the McClintock model for void growth.

4.2.1 Effect of pre-strain on stress-strain curve

The engineering stress-strain curves were obtained by dividing the load-displacement by the gauge length cross sectional area and gauge length respectively. However it is more meaningful to use true stress-true strain curves, as they will be used as input in the finite element simulations. Therefore all stress-strain curves in this study are plotted in term of true stress versus far field true strain where ϵ_{ff} is defined over the whole gauge length of the sample.

4.2.1.1 Tensile curve for the sample without voids

The typical true stress-true strain curve of aluminum alloy 5052 tensile coupon without holes is plotted in Figure 4-1.

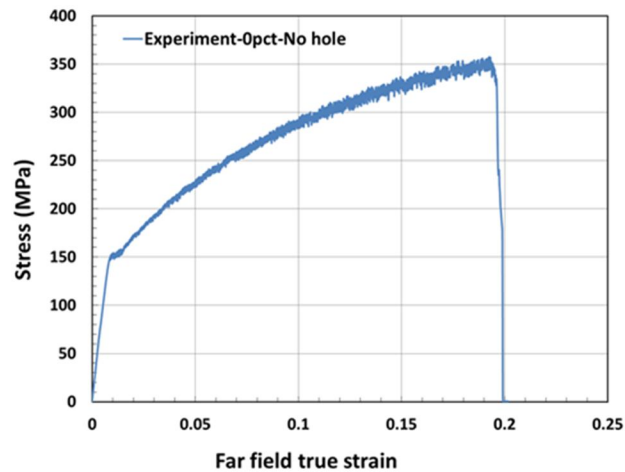


Figure 4-1 True stress versus far field true strain curve for the aluminum alloy 5052 without hole and pre-strain history

From Figure 4-1 one can see that this type of aluminum alloy (5052) has some specific mechanical behaviour. First the elastic region of the stress-strain curve is separated from the heterogeneous plastic flow region by a narrow portion of heterogeneous plastic deformation as shown in Figure 4-2. This heterogeneous flow is due to dislocation-solute atom interactions where there is a competition between diffused solute atoms pinning dislocations and dislocation prevailing over this obstacle (85). The sample is loaded elastically until point A where a local deformation band, also called Lüder band, forms and is responsible for the load drop to B. The Lüder band then propagates across the entire gage section (note that more

than one band may propagate simultaneously). Once the band has propagated to all parts of the gage length, heterogeneous deformation resumes.

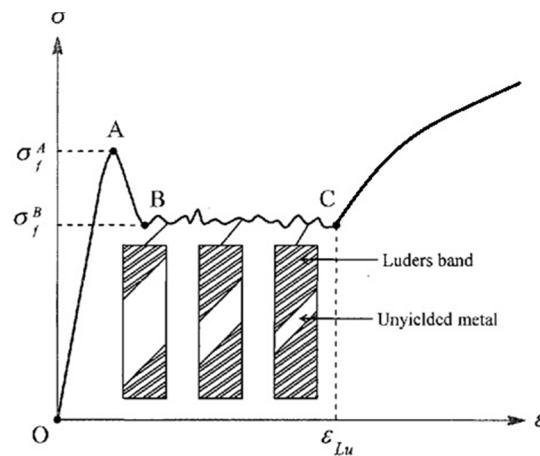


Figure 4-2: Typical stress-strain curve for Luders band; A=upper yield point B=lower yield point C=resume of heterogeneous deformation (86)

The second important tensile behaviour of aluminum alloy 5052 is the exhibition of a serration series superimposed on the parabolic portion of the curve. The load drops occur whenever the instantaneous strain rate in the specimen exceeds the rate of motions of the test machine cross-head. This high instantaneous strain rate is due to interactions between dislocations and solute atoms in the material also known as the Portevin-Le-Chatelier (PLC) effect (87). Figure 4-3 shows this load variation versus time for aluminum alloy 5052 which is studied by Wei Tong and Nian Zhang (88).

They discovered that the load serrations depend on the geometry and size of tensile samples and tensile machine stiffness. In addition to the Luder plateau and PLC behaviour; an ultimate tensile strength of 350 MPa, yield stress of 150 MPa and strain to failure of 0.2 can be extracted from Figure 4-1 for aluminum alloy 5052 tensile coupon without hole.

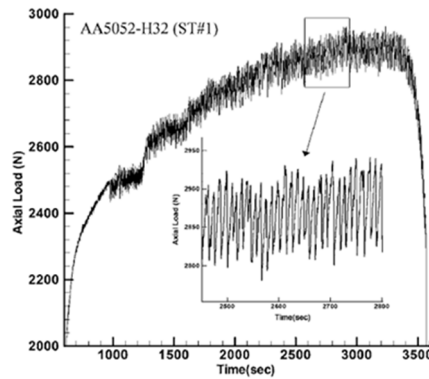


Figure 4-3: Uniaxial tensile serrated flow curves of AA5052-H32 standard test coupon ST1 (surface condition: as-received); originally recorded axial load versus time data $P(t)$ (88)

4.2.1.2 Tensile curves for the samples with 10 holes and different pre-strain histories

The true stress-true strain curves are plotted in Figure 4.4 (a) to (f) for 0%, 2.8%, 5.4%, 8.5%, 11.5%, 14.3% pre-strain histories respectively. One can see that the yield stress and ultimate tensile strength vary for different pre-strain histories. From Figure 4.4, it can be seen that the yield stress value increases by increasing the pre-strain histories. Also the ultimate tensile strength (UTS) is increasing with increasing pre-strain amounts. Differences in failure strains between flow curves with similar amounts of pre-strain is probably due to the fact that large intermetallic particles in aluminum alloy 5052 are not homogeneously distributed which could cause earlier or delayed failure. Furthermore, it will be shown that deformation bands occur randomly on the sample surface which could also affect when failure between voids occurs.

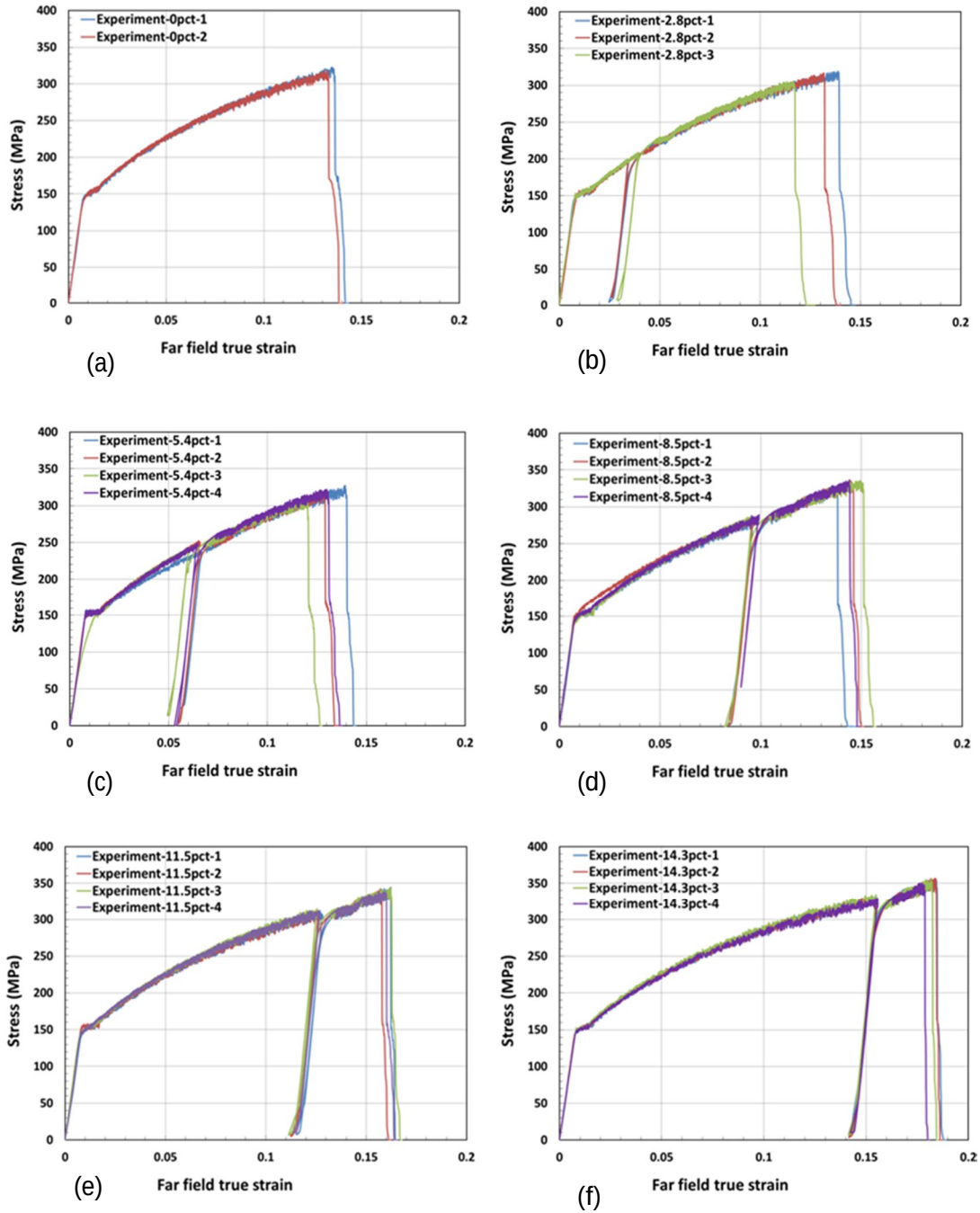


Figure 4-4: Plot of true stress versus far field true strain for aluminum alloy 5052 with different pre-strain histories; a) 0% PS, b) 2.8% PS, c) 5.4% PS, d) 8.5% PS, e) 11.5% PS, f) 14.3% PS

4.2.2 Effect of pre-strain on fracture surface area (ductility)

The ductility of a material can be expressed quantitatively using the reduction in area and is given by the true fracture strain ε_f which characterizes the degree of plastic deformation that has been sustained at fracture.

$$\varepsilon_f = \ln(A_0/A_f) \quad \text{Equation 4-1}$$

Where A_0 is the initial cross-sectional area and A_f is the cross-sectional area at fracture. A_0 is given by the initial ligament between the voids multiplied by the thickness of the sample. The area reduction of the samples was extracted from images of the fracture surface obtained from a scanning electron microscope (SEM) for different amounts of pre-strain, as shown in Figure 4-5. The same parameters for all pre-strain amounts were used to take the SEM images; i.e. a beam of 20 keV, a magnification of 1500x and care was taken to place the electron beam perpendicular to the fracture surface.

Changes in ductility (true fracture strain) with the amount of pre-strain are shown in Figure 4-6. Nine measurements were taken corresponding to the nine inter-hole spaces and the standard deviation of these nine measurements is represented by the error bars in Figure 4-6.

A decrease in true strain to fracture is observed with increasing pre-strain (error bars in Figure 4-6). Although the ductility of the material decreases with increasing pre-strain, these variations are within the experimental scatter given by the error bars.

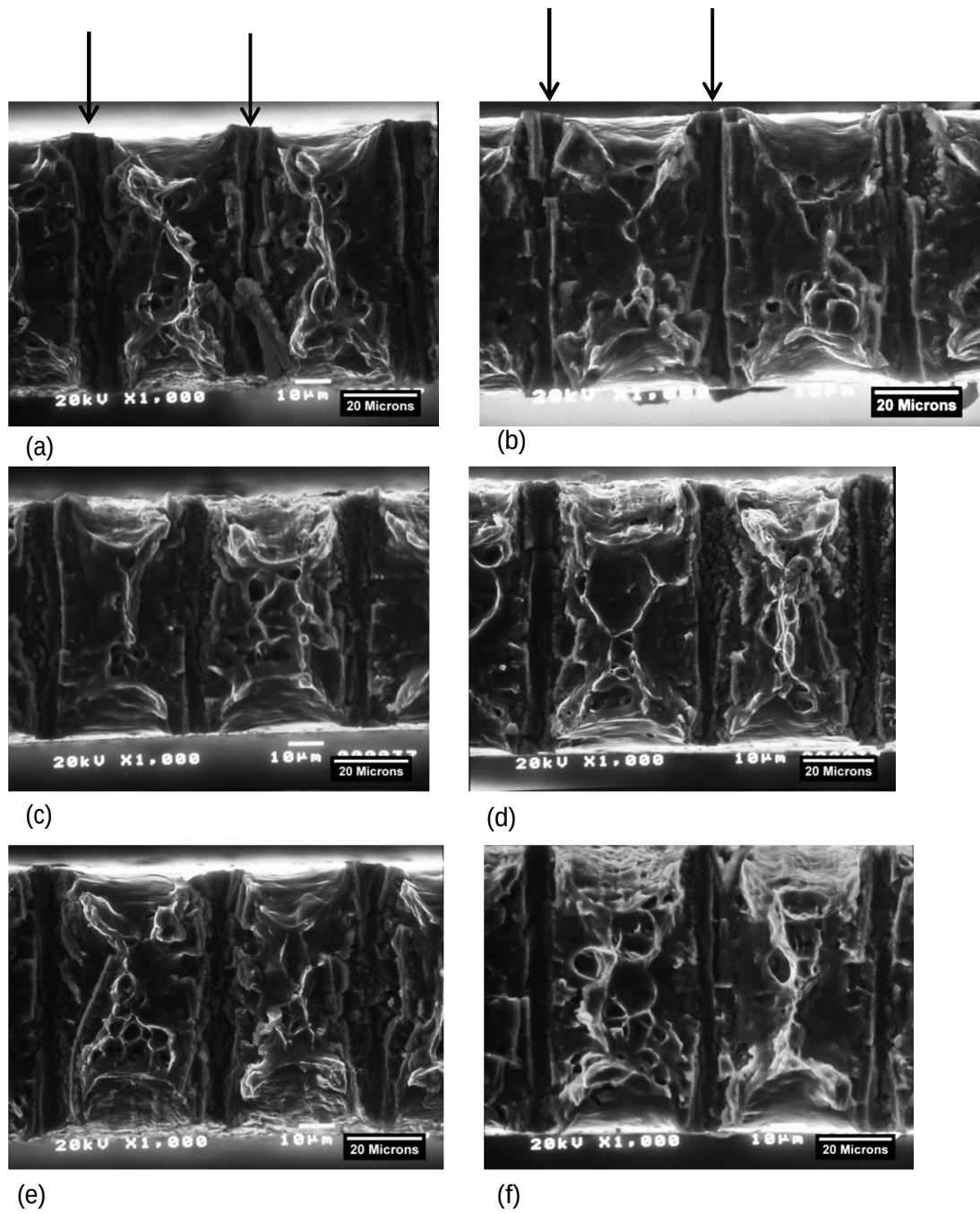


Figure 4-5: SEM images of the area extracted from fracture surfaces of aluminum alloy 5052 tensile sample for different pre-strain histories; a=0%, b=2.8%, c=5.4%, d=8.5%, e=11.5%, f= 14.3%. The black arrows indicate the location of laser holes.

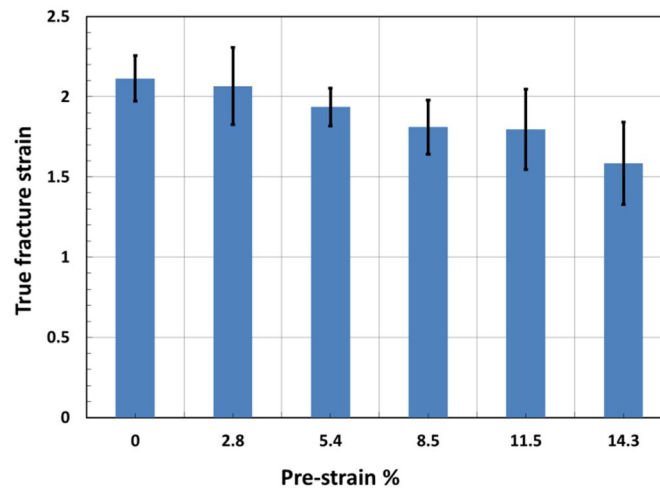


Figure 4-6: Effect of pre-strain on ductility; reduction in area for different pre-strain histories; error bars are based on standard deviation

4.2.3 Effect of pre-strain on void growth and coalescence

The effect of pre-strain on void growth and coalescence was studied by recording the evolution of the major void diameter with pre-strain. For this purpose the same initial void geometry for all pre-strain histories was considered and tensile samples with different pre-strain histories were pulled in the same direction under an optical microscope with a magnification of 20x.

Figure 4-7 shows typical optical microscope images of the sample in the initial state, during void growth, at the onset of coalescence, failure of the ligament between holes and final failure of the sample. From Figure 4-7 (b) to (d), one can see how the coalescence happens by internal necking of the ligament between the holes which were oriented at 90° with respect to the tensile axis (39).

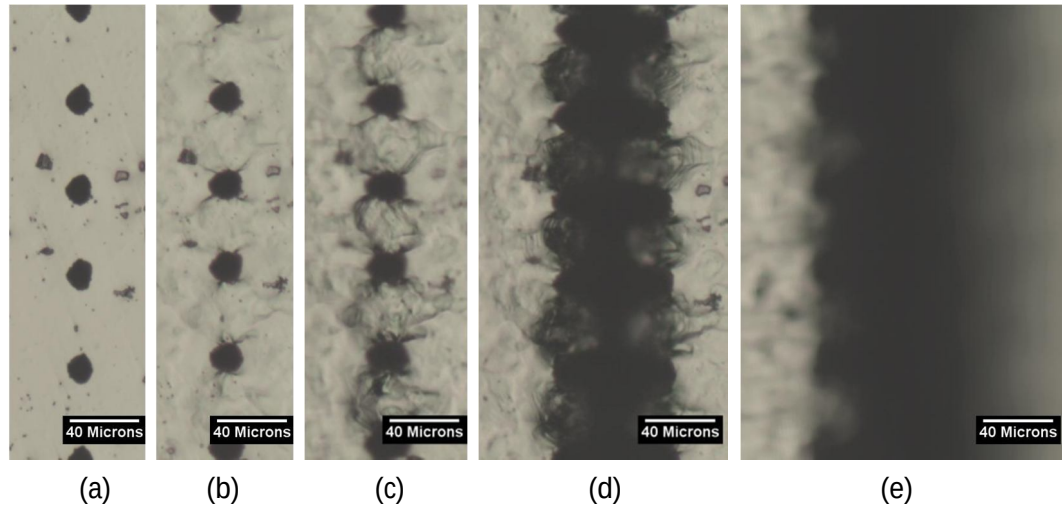


Figure 4-7: Optical images showing the void growth and coalescence in a line of holes oriented at 90° with respect to the horizontal tensile axis; a) beginning of the deformation b) while growing c) onset of coalescence d) failure at holes e) final failure in edge of tensile sample

4.2.3.1 Effect of pre-strain on major void diameter growth

From the optical pictures taken in-situ during the tensile tests, the major diameter of the void or void length a , is measured for 0%, 2.8%, 5.4%, 8.5%, 11.5% and 14.3% pre-strain histories. Figure 3-11 shows a schematic diagram of the void dimensions. The major void growth evolution is defined as a/a_0 where a_0 is the initial major diameter and is plotted as a function of the far field true strain ϵ_{ff} defined as $\ln(e_{ff}+1)$, where e_{ff} is the far field engineering strain defined over the gauge length of the pre-strained sample. Figure 4-9 (a) to (f) shows the void growth evolution for aluminum alloy 5052 samples with different pre-strain histories.

Figure 4-10 shows that by increasing the amount of pre-strain from 0% to 14.3%, voids grow more rapidly and the sample fails at a smaller strain. Also for all different pre-strain histories a stepwise void growth evolution is observed. Void growth is constant for given amount of applied strain and suddenly jumps to the next void growth level; this behaviour continues until final failure.

4.2.3.2 Effect of pre-strain on void coalescence and failure

Void coalescence is the last step of the ductile fracture process. Normally the onset of void coalescence is related to the instant where a relatively homogenous deformation localizes in the horizontal inter-void ligament (61). It is difficult to experimentally observe void coalescence and only the final failure strains were extracted, assuming that from coalescence to failure there is only a negligible amount of far field strain.

Strains to failure were extracted from the stress-strain curves and are plotted as a function of the amount of pre-strain in Figure 4-10. From Figure 4-11 it can be seen that a pre-strain history significantly reduces void growth and coalescence strain and this decrease is linear within the error bars.

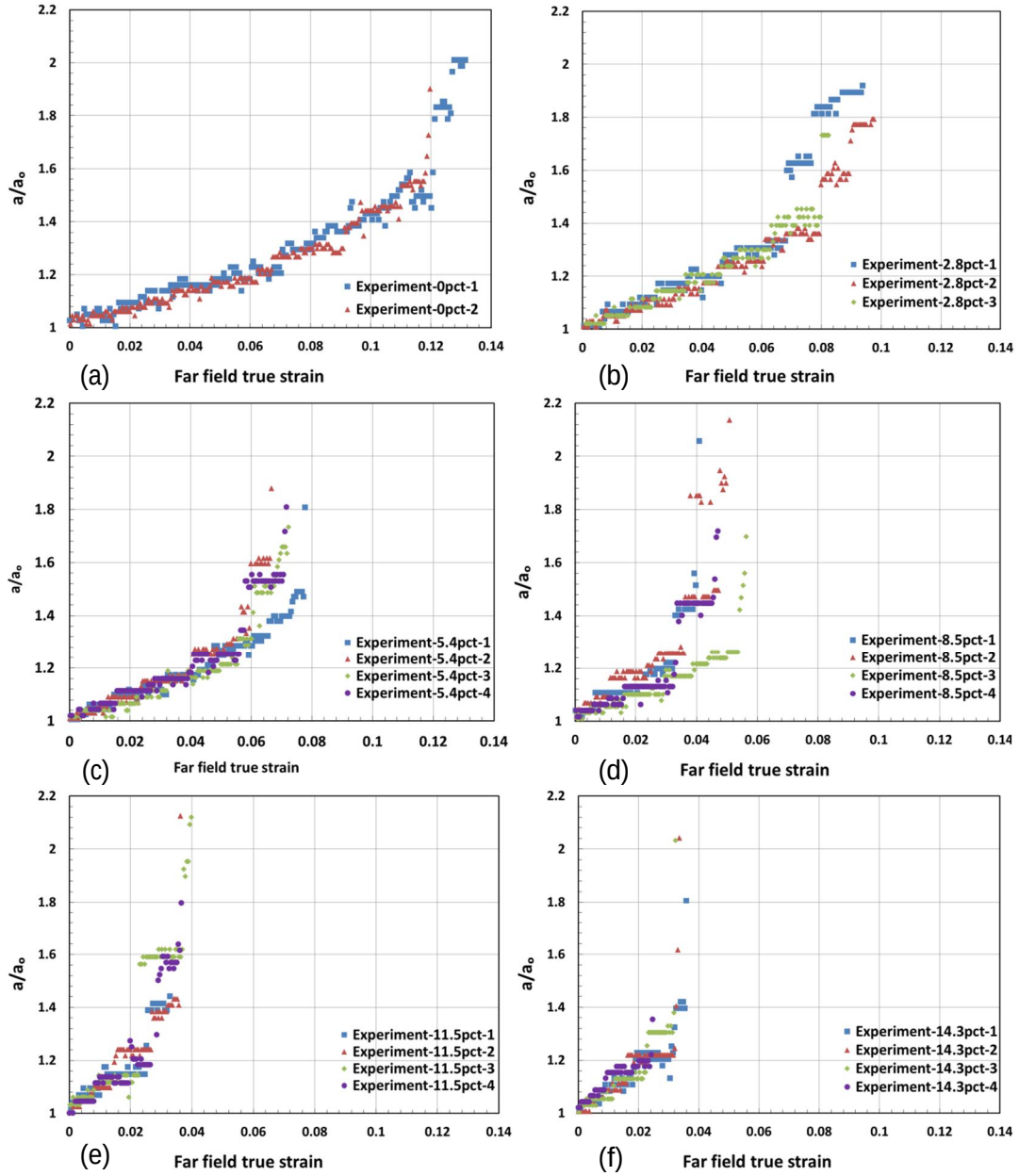


Figure 4-8: Evolution of the normalized major void diameter (a/a_0) versus far field true strain for holes drilled in a 5052 aluminum alloy with different pre-strain histories; a) 0% PS, b) 2.8% PS, c) 5.4% PS, d) 8.5% PS, e) 11.5% PS, f) 14.3% PS

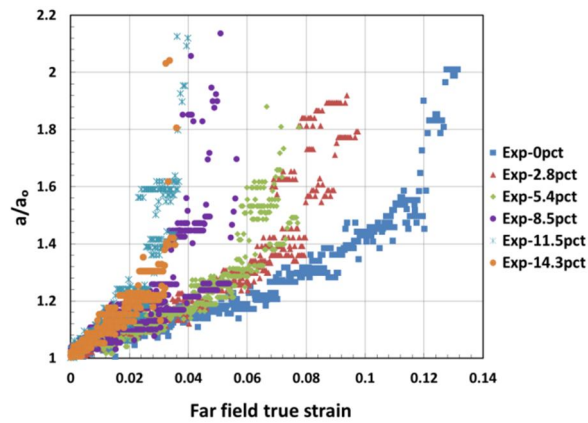


Figure 4-9: Evolution of the normalized major void diameter (a/a_0) versus far field true strain for holes drilled in a 5052 aluminum alloy with different pre-strain histories; comparative curve including all experimental data series for each pre-strain amount

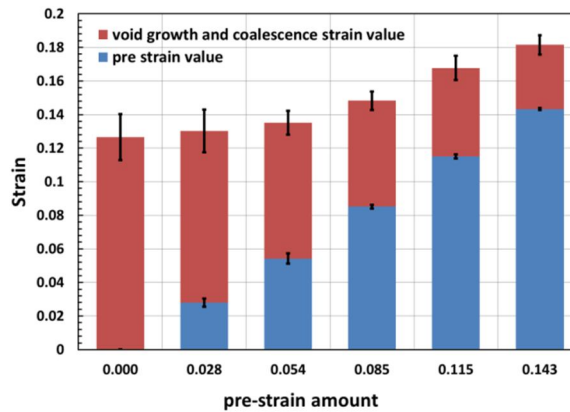


Figure 4-10: Plot of void growth and coalescence strain for different pre-strain value; error bars are based on standard deviation

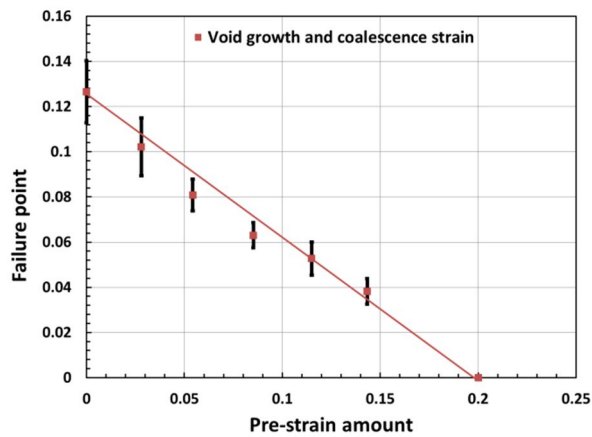


Figure 4-11: Plot of strain to failure versus pre-strain amount for pure pre-strain test. Error bars are based on standard deviation. In the x-axis, value of 0.2 indicates the max strain to failure for the sample without pre-strain history

4.2.4 Discussion

4.2.4.1 Behaviour of stress-strain curve

It was observed that with increasing pre-strain, the yield point increases and the work hardening capacity of the material decreases. This is explained by the fact that pre-strain induces strain hardening which causes a dramatic increase in the number of dislocation-dislocation interactions which reduces dislocation movement. Since the dislocation density increases with plastic deformation, the material is hardened and a higher yield stress is required to resume plastic deformation. Therefore the material with higher pre-strain value is more hardened and has a higher yield stress when reloading. This increase in yield stress and ultimate tensile strength have been seen in the previous study on API 5L X65 pipe which was shown in the literature review (47). However, with higher pre-strain, the work hardening capacity of the material decreases which will affect fracture as explained in the next section.

4.2.4.2 Change in fracture surface area

Figure 4-6 shows that the ductility of the sample decreased with increasing pre-strain within large error bars, but only slightly compared to the large decrease in far field true strain at failure seen Figure 4-10. Also from Figure 4-5 (a) to (f) there is no clear trend in fracture surface morphology where both secondary void nucleation and necking down of the intervoid ligament to a line are observed. These results suggest that failure is mainly controlled by strain localization and not by ductile fracture.

The most common type of ductile fracture mode is presented in Figure 4-12, where this process occurs in several stages; first small cavities are formed in the interior of the cross section (void nucleation), then these cavities grow (void growth) until a critical condition is reached at which point they link with each other (void coalescence).

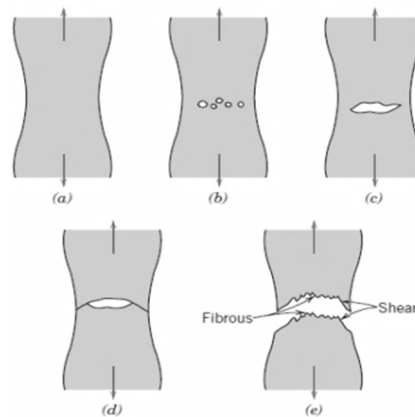


Figure 4-12: Schematic diagram of ductile fracture. (a) Initial necking. (b) Small cavity formation. (c) Coalescence of cavities to form a crack. (d) Crack propagation. (e) Final shear fracture at a 45 degrees angle relative to the tensile direction (75)

Plastic deformation of void-containing material demonstrates two distinct phases, the homogeneous phase and localized phase. In fact, there is competition between two modes of deformation, localized necking and homogeneous plastic deformation (13).

Before void coalescence, the plastic deformation develops in a dilatational and macroscopically homogeneous way while the plane stress condition is achieved. At the onset of coalescence the homogeneous phase terminates and shifts to a localized plastic deformation where the plane strain condition is achieved. So all the deformation acts in the vertical direction and through thickness, with no additional

plastic deformation in tensile direction; which causes the fracture surface to neck down to a point, see Figure 4-13.

From our experimental observations it can be suggested that pre-strain affects strain localization due to the lower work hardening capability of the matrix and thus plays a major role in driving the fracture process in hole-containing pre-strained material.



Figure 4-13: Ductile fracture mode by 100% reduction in area

4.2.4.3 Major void diameter growth and coalescence

Void growth and coalescence analysis was carried out in terms of the normalized major void diameter (a/a_0) evolution versus far field true strain and strain to failure for different pre-strain histories.

The experimental results revealed that void growth accelerates with increasing pre-strain. This behavior can be rationalized again in terms of the work hardening capacity in the samples. At low pre-strains the material's work hardening rate is still high which delays void growth. When pre-strain increases the work hardening capacity of the material decreases and void growth can accelerate,

without being slowed down by regions undergoing hardening. From a previous study on HSLA-80, it is interesting to note that at lower pre-strain values (less than 2%) the work hardening remains constant due to the yield propagation effect which results in no hardening on pre-straining (41).

The effect of work hardening rate on void growth has already been presented by Thomason in his review on ductile fracture (13) where he compared theoretical void growth in non-hardening and linear hardening materials. Our experimental results confirm these theoretical findings and place them in the context of the effect of pre-strain on ductile fracture.

Figure 4-10 shows that larger pre-strain histories significantly decrease the void coalescence strain. In fact the higher pre-strained material tends to fail earlier compared to lower pre-strained samples, because of dislocation accumulation which reduces the dislocation movement and lowers the work hardening rate which in turns promotes strain localization.

From the Figure 4-9 for different pre-strain histories, one can see that the evolution of the normalized void diameter is constant for some range of true strain, and then it suddenly jumps to the next step where this behaviour repeats until failure. This gradual increase in void growth following a stepwise behaviour is due to shear band propagation through gauge length of the sample. Shear bands propagate randomly along the gauge length of aluminum alloy 5052 and these sudden jumps happen whenever shear bands appears in the region of the voids.

4.2.4.3.1 Comparison between void growth and McClintock Model in the literature

The McClintock model for cylindrical void growth was described in section 2.2.2.2.1; its main governing equation was shown in equation 2-2. The comparison between McClintock void growth model and experiment for aluminum alloy 5052 was studied by A. Weck et al. (39) and the model predicts relatively well void growth at lower far field true strain for 0% pre-strain tests.

However, the strain hardening has a strong effect on local strain, $\ln(a/a_0)$ and the hardening exponent has to be changed accordingly in the McClintock model to account for the different pre-strain histories. The strain hardening component, n has been determined from experimental stress-strain curves and is equal to 0.32, 0.27, 0.24, 0.19, 0.16 and 0.11 for 0%, 2.8%, 5.4%, 8.5%, 11.5% and 14.3% pre-strain respectively.

For uniaxial tension, $\sigma_b=0$, $\sigma_z=0$, $\sigma_a=Y$, $\varepsilon_z=-1/2\varepsilon_a$, $\varepsilon_b=-1/2\varepsilon_a$ and $\varepsilon_a=\bar{\varepsilon}$ where Y and $\bar{\varepsilon}$ are von mises stress and far field true strain respectively. Also, the main McClintock model is based on mean radius R (equation 4-2) and it needs to be separated for both major and minor diameter by using the eccentricity factor, m (28) which is defined from Equation 4-3.

$$R = \frac{a+b}{2} \quad \text{Equation 4-2}$$

$$m = \frac{a-b}{a+b} = \frac{\sigma_a-\sigma_b}{\sigma_a+\sigma_b} + \left(m^0 - \frac{\sigma_a-\sigma_b}{\sigma_a+\sigma_b} \right) \exp \left[-\frac{\sqrt{3}\bar{\varepsilon}}{(1-n)} \sinh \left(\frac{\sqrt{3}(1-n)}{2} \frac{\sigma_a+\sigma_b}{Y} \right) \right] \quad \text{Equation 4-3}$$

Finally, the main McClintock equation for mean radius reduces equation 2-2 for major diameter to:

$$\frac{a}{a_0} = \exp \left[\varepsilon_a \left(\frac{\sqrt{3}}{2(1-n)} \sinh \left(\frac{\sqrt{3}(1-n)}{2} \right) + \frac{1}{4} \right) \right] \left[2 - \exp \left[\frac{-\sqrt{3}\varepsilon_a}{(1-n)} \sinh \left(\frac{\sqrt{3}(1-n)}{2} \right) \right] \right] \quad \text{Equation 4-4}$$

Figure 4-14 shows the variation of McClintock model. Although, a difference in McClintock model has been seen for various hardened material, but only slightly compare to the large change in void growth rate. Therefore, the current McClintock model is not able to predict void growth and coalescence in pre-hardened material and needs further improvement see Figure 4-15.

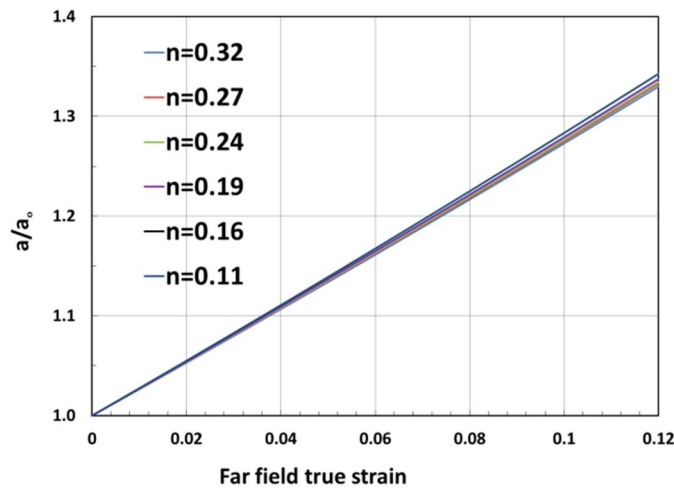


Figure 4-14 Variation of McClintock model for maximum and minimum strain hardening component; $n=0.32$ and $n=0.11$

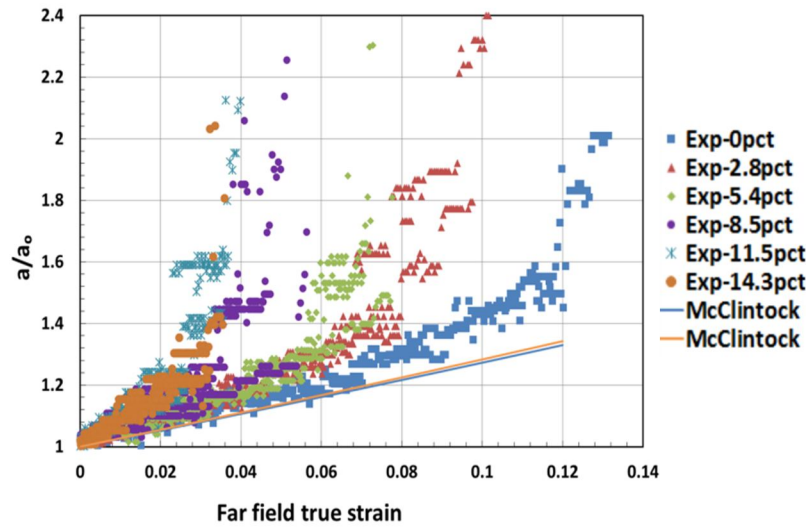


Figure 4-15: Comparison between the McClintock model prediction for void growth and the experimental results

4.3 Effect of strain path changes on ductile fracture

As shown in the literature, plastic deformation does not always occur in the same direction and metals can undergo complex loading sequences. These complex strain paths lead to a reorganization of the dislocation structure and formation of cell block boundaries (CBBs). Most of the research in this area was done either on microstructure and texture evolution after changing the strain path, or on mechanical and tensile behaviour of the samples under complex loading sequences. Therefore more quantitative experimental data at smaller length-scales are required to study the effect of strain path changes on ductile fracture.

In the first part of this section, the effect of strain path changes on the stress-strain curve is presented and a strong dependency of the yield point on strain path changes is shown; while the second part deals with the effect of strain path changes on digital image correlation maps and local strain evolution with respect to plastic deformation. Finally the effect of strain path changes on void growth and coalescence in aluminum alloy 5052 is presented.

4.3.1 Effect of strain path changes on stress-strain curve

Normally the evolution of mechanical properties, in particular the behaviour of yield stress under complex loading sequences is one of the most challenging exercises in material science. Although there are numerous modeling approaches for understanding the hardening or softening effects after strain path changes, more experimental data are still required to validate these models.

In this study, the evolution of yield stress after strain path change with 90° angle with respect to the pre-strain direction is verified for aluminum alloy 5052.

The true stress versus far field true strain curves for 0%, 4.3%, 8%, 12.6% and 16.7% pre-strain histories are plotted in Figure 4-16 (a) to (d) respectively. From these curves, one can see that the yield stress drops for all pre-strain histories after strain path changes and that yielding is much more progressive than during the pre-straining part. Difference in failure strains between the flow curves for the same pre-strain amount is probably due to the randomness of shear band evolution along the gauge length of the samples as discussed previously.

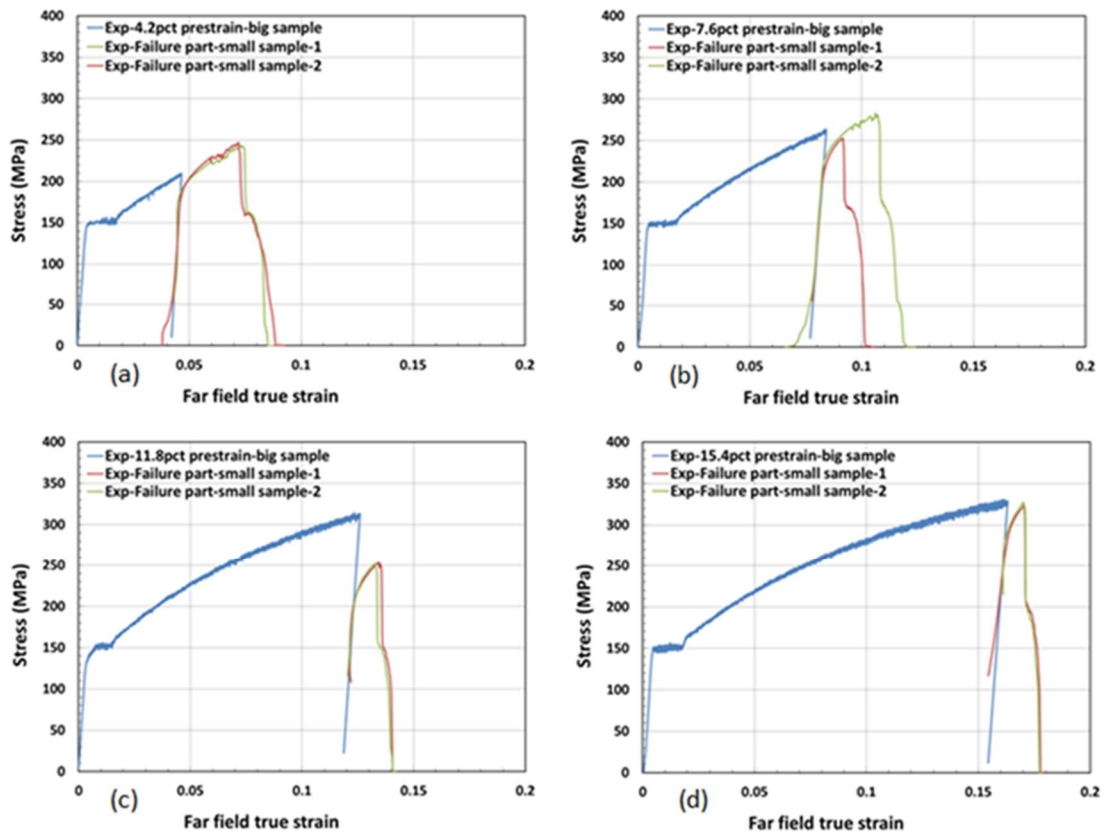


Figure 4-16: Plot of true stress versus far field true strain for aluminum alloy 5052 after strain path changes with 90 degrees respect to pre-strain part; a) 4.2% PS, b) 7.96% PS, c) 11.8% PS and d) 15.4% PS

4.3.2 Effect of strain path changes on fracture surface area

The same parameter as pure pre-strain test for SEM was used to extract the area reduction of the samples for different amounts of pre-strain, as shown in Figure 4-17. Nine measurements were taken corresponding to the nine inter-hole spaces and the standard deviation of these nine measurements is represented by the error bars in Figure 4-18. A decrease in true strain to fracture is observed with increasing pre-strain. Although the ductility of the material decreases with increasing pre-strain, these variations are within the experimental scatter given by the error bars.

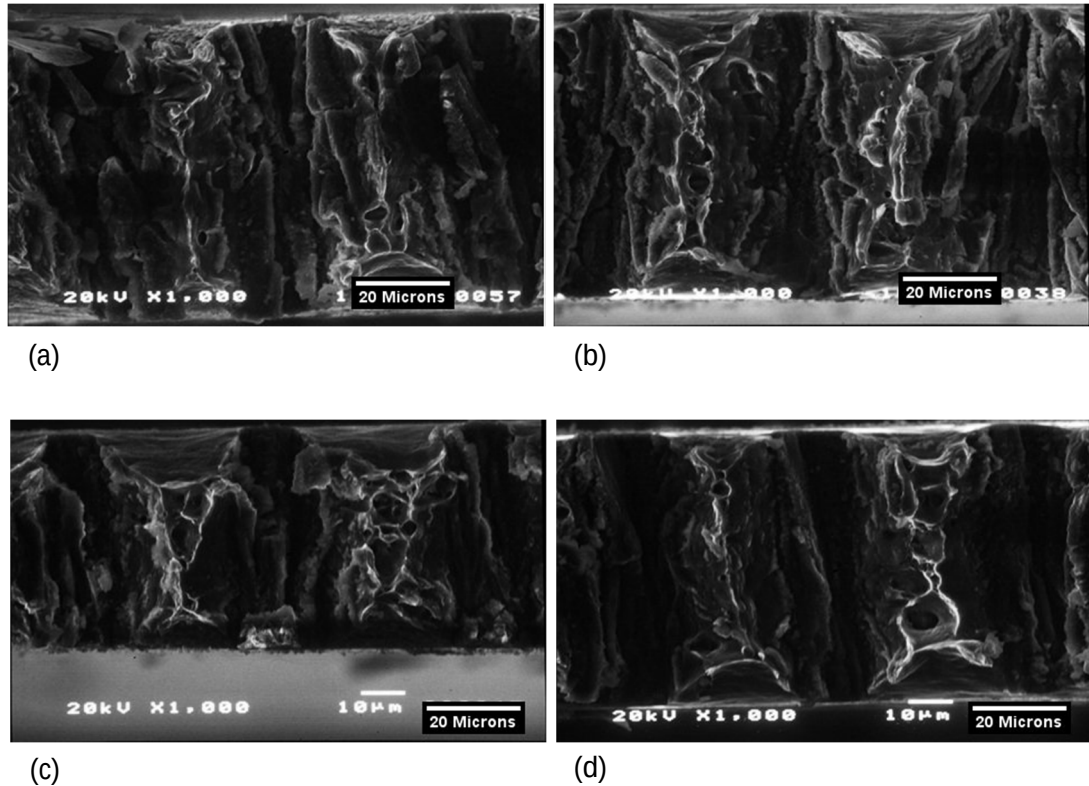


Figure 4-17 SEM images of the area extracted from fracture surfaces of aluminum alloy 5052 tensile sample for strain path changes tests; a=4.2% PS, b=7.96% PS, c=11.8% PS, d=15.4% PS.

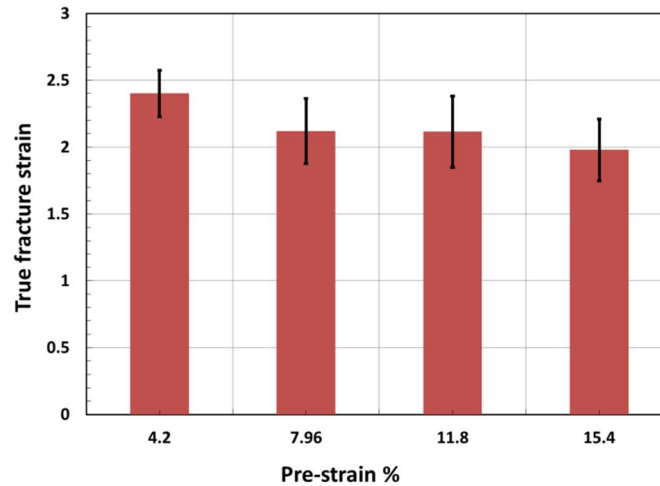


Figure 4-18 Effect of strain path changes on ductility; reduction in area for different pre-strain histories; error bars are based on standard deviation

4.3.3 Evolution of digital image correlation (DIC) map under strain path changes

The digital image correlation (DIC) technique is used for accurate 2D and 3D measurements of changes in an image. This method is often applied to obtain deformation (engineering), displacement, and strain data. In this study DIC has been utilized to obtain a full 2-D strain maps. The first part of this section shows how the exact amount of pre-strain can be obtained from DIC while the second part deals with the evolution of the local strain between voids after strain path changes. Determination of the local strain between voids is an important factor to see strain concentration between and around the holes and to potentially extract local strain for failure models.

4.3.3.1 Obtaining the precise pre-strain amount for small tensile samples; using DIC map

The digital image correlation images are shown in Figure 4.18 and show the strain distribution along the gauge length of big tensile sample of aluminum alloy

5052. In Figure 4.18 (a) to (f), the three main steps of tensile behaviour in aluminum alloy 5052 can be observed, namely the elastic region, Luders band propagation and heterogeneous plastic deformation. The final 2D strain field for different pre-strained aluminum samples is plotted in Figure 4-20 (a) to (d). The color intensity along the y axis from Figure 4-20 shows the pre-strain amount and its difference with nominal pre-strain amount which was obtained from tensile machine recording.

4.3.3.2 Effect of strain path changes on local strains between voids

Normally the strain distribution is not constant throughout the whole gauge of the tensile sample and is highly dependent on the shape and geometry of the sample. It is expected that stress concentration due to the voids will result in strain localization between voids. Digital image correlation maps are presented in Figure 4-21 and Figure 4-22 showing the local strain between voids for different amounts of pre-strain. The changes in color intensity in Figure 4-21 (c), (f) and Figure 4-22 (c), (f) represent the strain in the sample from initial state to the onset of coalescence. For the 12.58 % pre-strain case, because of the blurry pattern at the bottom half of the sample, image processing was carried out only for the area in focus. Furthermore, for better understanding the effect of strain path changes on local strain, the evolution of local strain across the line of holes is presented in Figure 4-23 for different pre-strained tensile coupons.

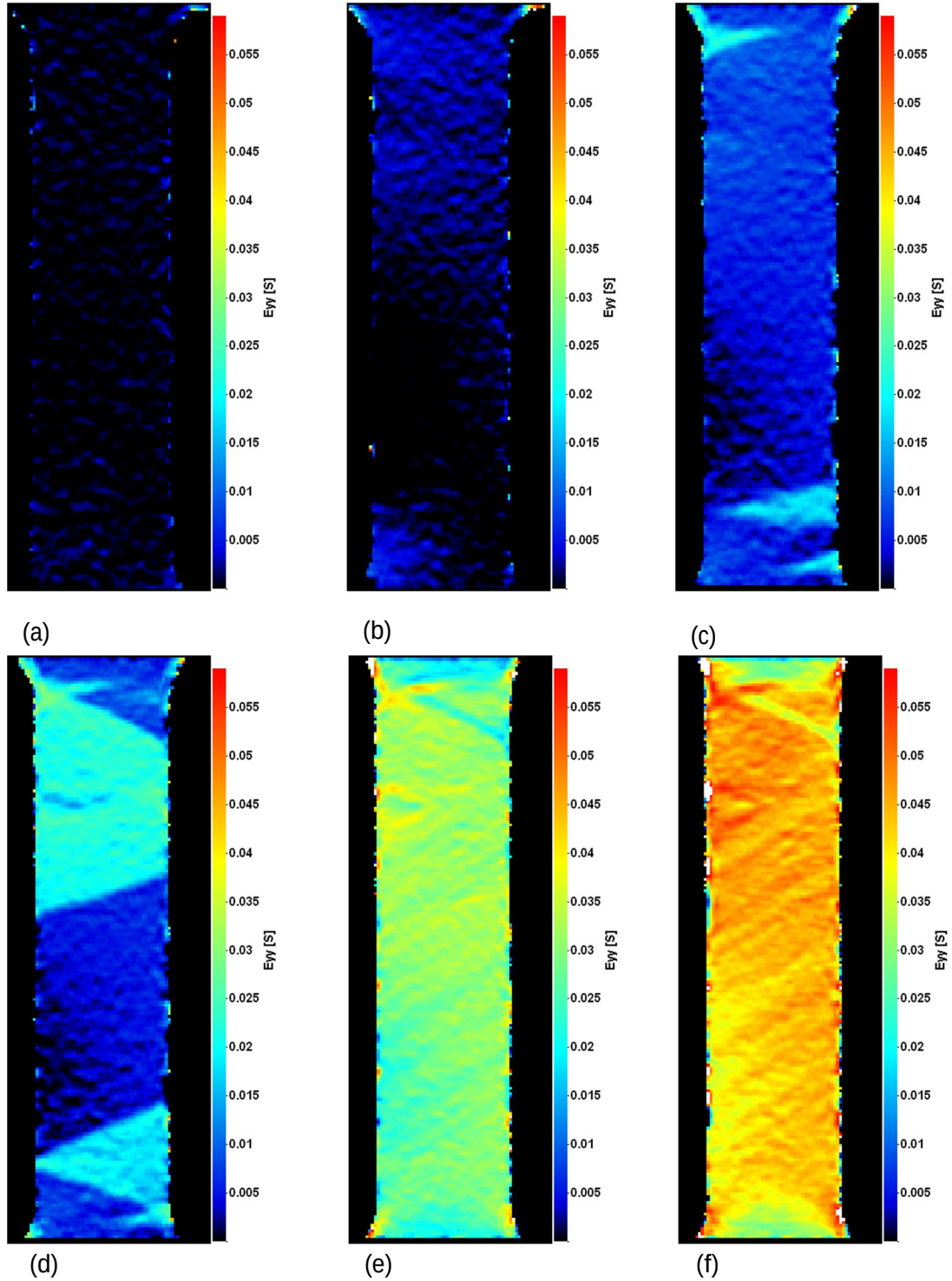


Figure 4-19: Digital image correlation maps for the big tensile sample with 4.6% nominal pre-strain history; a and b) the elastic deformation part, c and d) propagation of Luders band, e and f) heterogeneous plastic deformation where shear band forms and propagate along gauge length randomly

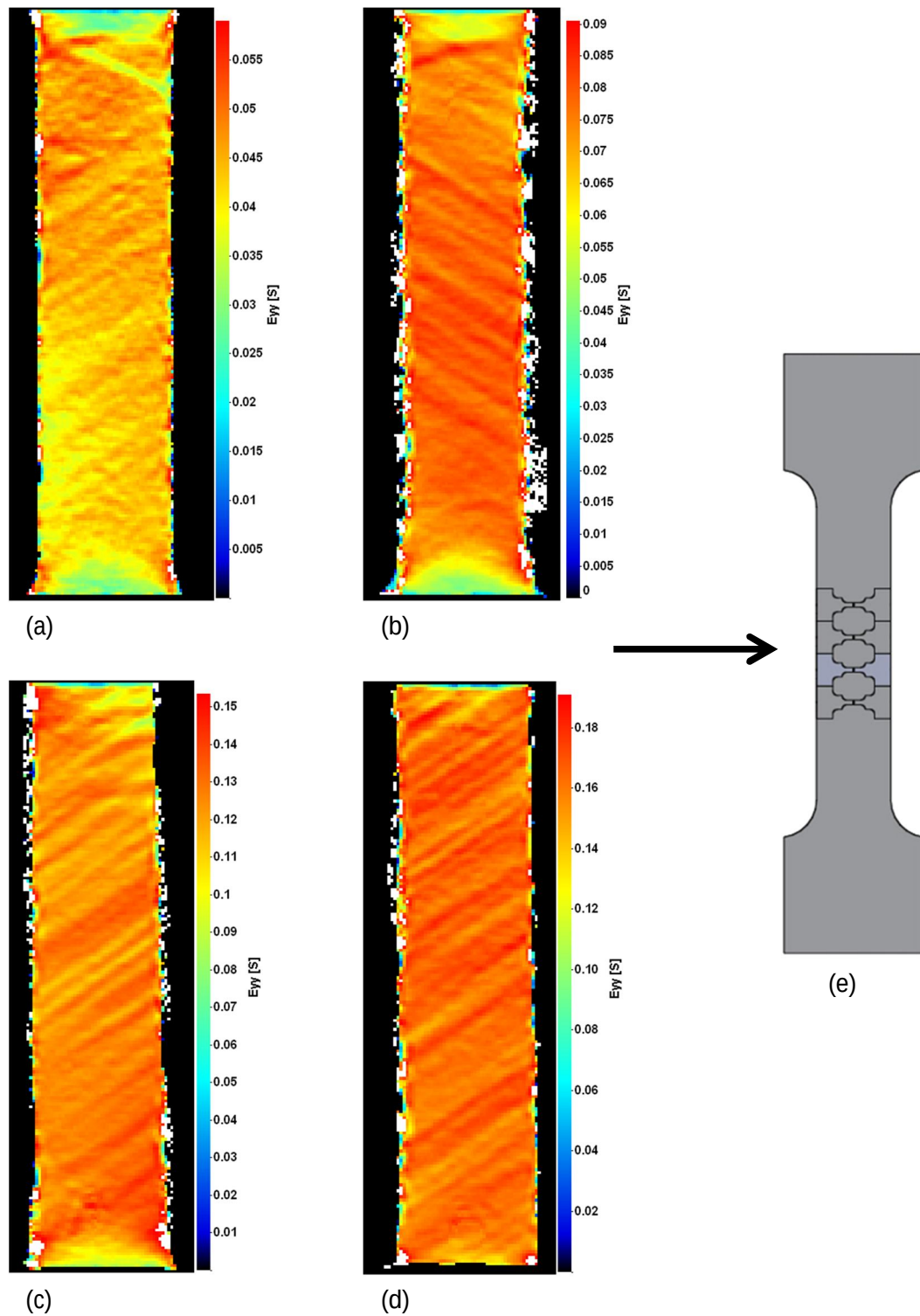


Figure 4-20: Digital image correlation maps for different pre-strained big tensile coupon; a) 5% nominal pre-strain, b) 10% nominal pre-strain, c) 17.8% nominal pre-strain, d) 21.42% nominal pre-strain and e) schematic diagram of big tensile sample including extracted small samples

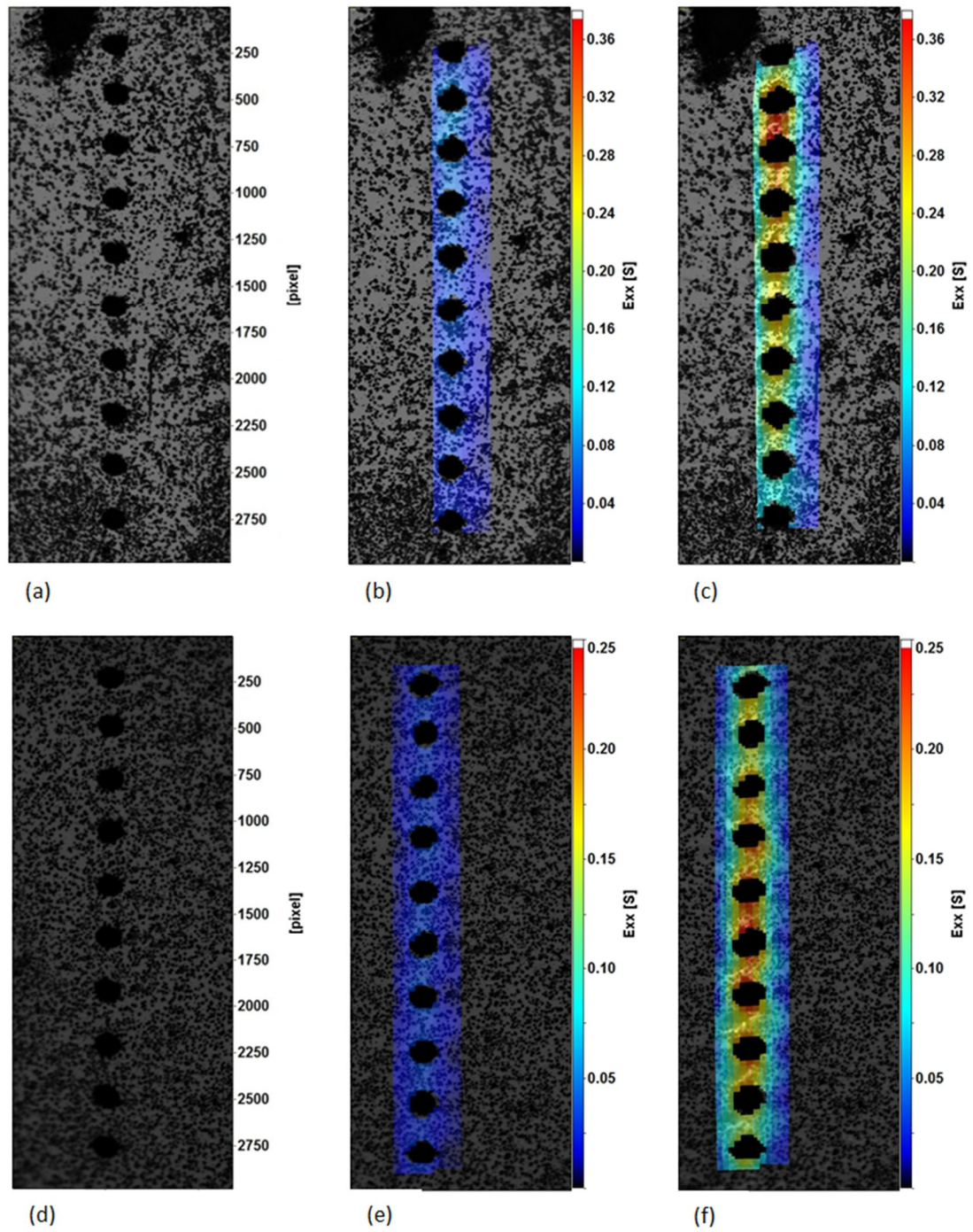


Figure 4-21: 2D strain field DIC images for different pre-strained sample showing the local strain between the voids in raw images, while deforming and onset of coalescence; a, b, c: 4.27% pre-strained sample, d, e, f: 7.96% pre-strained sample

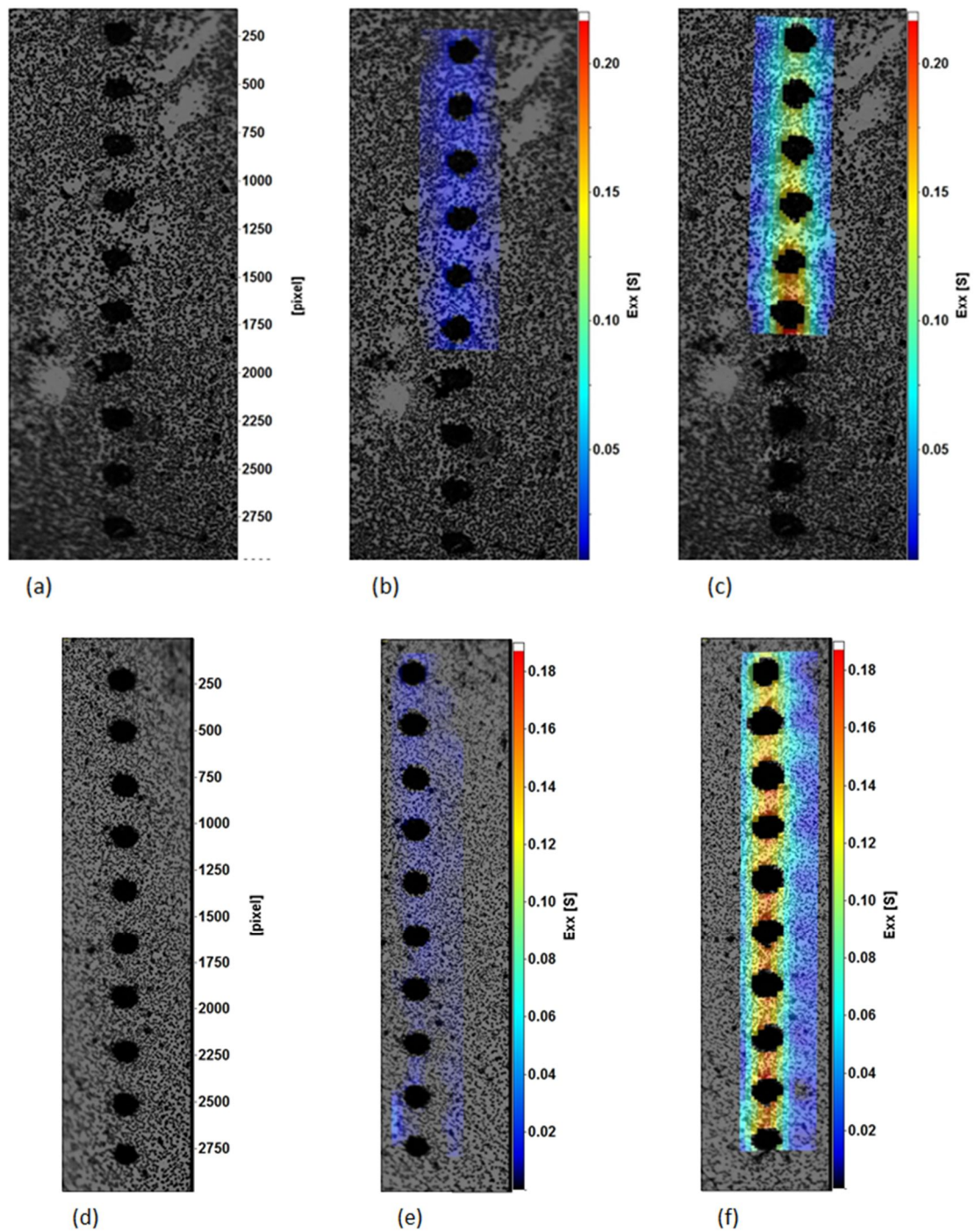


Figure 4-22: 2D strain field DIC images for different pre-strained sample showing the local strain between the voids in raw images, while deforming and onset of coalescence; a, b, c: 12.58% pre-strained sample (half of the sample was analyzed), d, e, f: 16.70% pre-strained sample

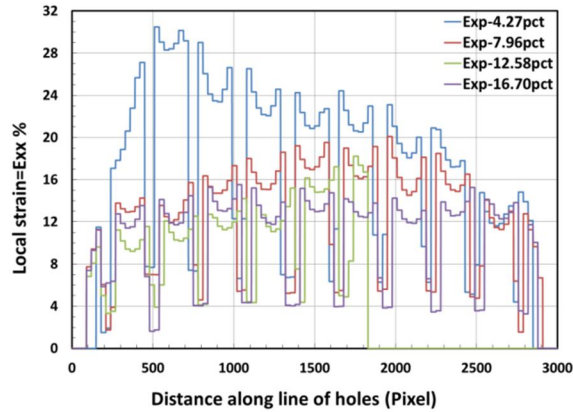


Figure 4-23: Plot of evolution of local strain versus line of the holes for different pre-strained tensile sample. In the case of 12.58 half of the sample was analyzed; curves were extracted from Davis software

4.3.4 Effect of strain path changes on void growth and coalescence

The effect of strain path changes on void growth and coalescence was investigated in terms of major void diameter evolution and failure strain amounts for different pre-strained tensile coupons. The same initial void geometry for all tensile pre-strained samples was obtained by laser micromachining 10 holes in the gauge of the sample, and the samples were pulled at an angle of 90° with respect to the pre-strain direction. All tests were carried out under an optical microscope with 20x magnification. The change in strain path during reloading has a strong effect on the major void growth diameter and strain to failure value which is explained in the following sections.

4.3.4.1 Effect of strain path changes on major void diameter growth

Figure 4-24 shows the effect of strain path changes on normalized major void diameter for different pre-strained tensile samples. Void growth evolution is very similar to the behaviour seen in the previous sections on pure pre-strain, where voids grow more rapidly at higher pre-strain and the sample fails at a smaller strain.

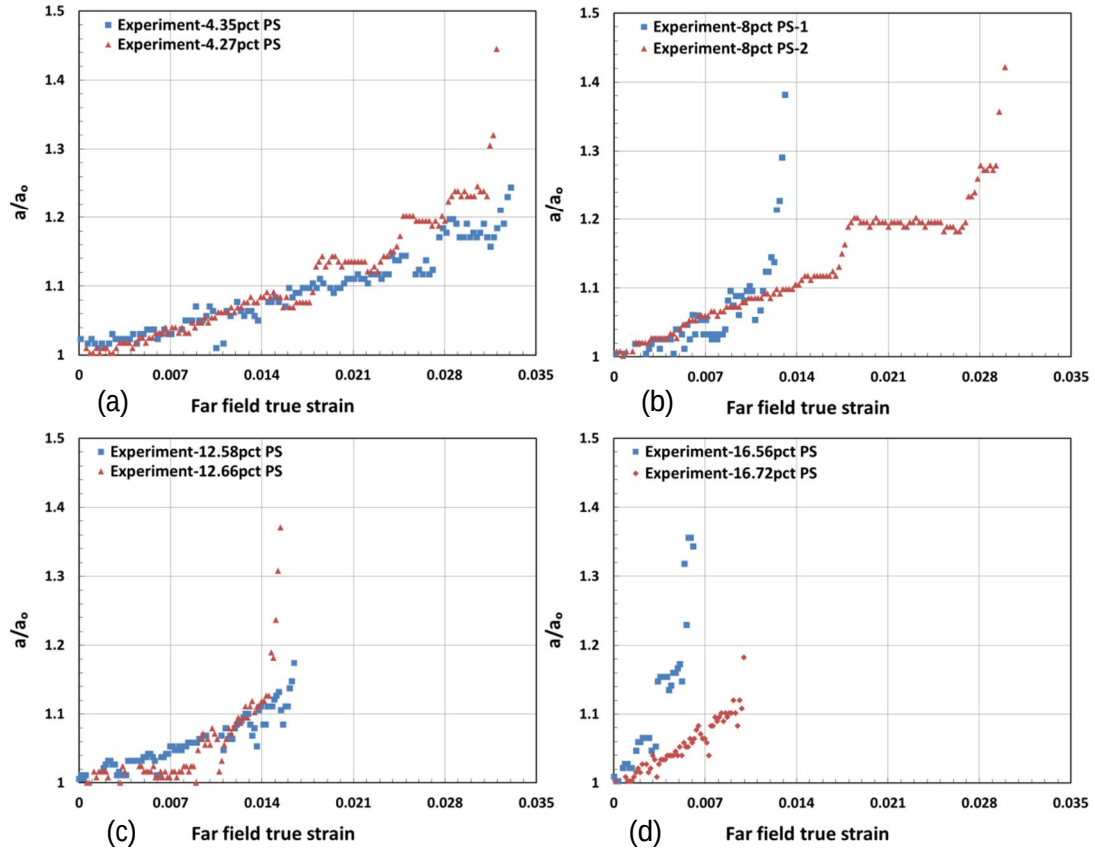


Figure 4-24: Evolution of the normalized major void diameter (a/a_0) versus far field true strain for holes drilled in a 5052 aluminum alloy, after strain path changes; a) 4.3% PS, b) 8% PS, c) 12.62% PS and d) 16.64% PS

4.3.4.2 Effect of strain path changes on coalescence and failure

In order to obtain the effect of strain path changes on void coalescence and failure, strains to failure were extracted from the stress-strain curves and are plotted as a function of pre-strain amount in Figure 4-25. From Figure 4-25 one can see that strain-path change decreases void growth and coalescence strain significantly in comparison to pure pre strain tests (Figure 4-10), and that the samples with higher pre-strain failed earlier.

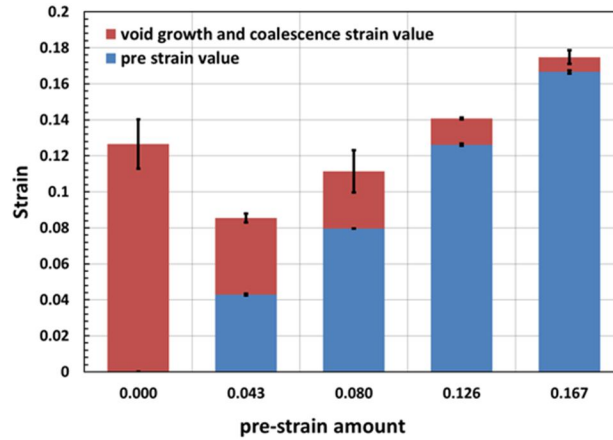


Figure 4-25: Plot of void growth and coalescence strain after strain path changes with 90 degrees with respect to pre-strain part; error bars are bases on standard deviation

4.3.5 Discussion

4.3.5.1 Behaviour of stress-strain curve

Experimental result shows a significant reduction in yield stress after strain path changes 90 degrees with respect to the pre-loading direction (Figure 4-16). This drop in yield stress has been extensively studied in the literature for different materials in using both modeling and experiments.

Normally strain path changes lead to a reorganization of the dislocation structure which generally results in plastic anisotropy, and a change in tensile behaviour, either softening or hardening (69) (70). The reduction or increase in work hardening rate happens when strain path changes results in either the activation of new slip systems or more difficult dislocation movement (74).

The drop in yield stress can be explained in terms of a softening effect under strain path changes. The Taylor model (72) for cell block boundaries (CBBs) revealed that under strain path changes, dislocations possess a mechanical charge (sign of burger vectors) causing a polarization of the CBBs. This polarization results

in the slip resistance being asymmetrical producing this drop in yield stress. Furthermore it was found that strain incompatibilities between grains which induce internal stresses or transient stagnation can also be the reason for this drop in yield stress (73). Finally local back stresses (from unloading part) may be present in the material, which help the dislocation mobility in the new direction. Therefore the dislocations can move more easily in the new direction which results in yield stress reduction during reloading.

There is a trend between pre-strain and yield stress where a higher pre-strain value results in a higher yield stress. The explanations drawn for the pure-pre-strain tests in section 4.2.4.1 are valid here too. The unexpected behaviour in the case of 11.8 % is due to the polishing step where the sample thickness was reduced significantly. For better prediction and extracting the model parameters, more quantitative experimental results at smaller length scales are needed.

4.3.5.2 Change in fracture surface area

Strain path changes results in Figure 4.17 shows the same behaviour as pure pre-strain tests where the true fracture strain (ductility) decreased with increasing pre-strain within large error bars, but only slightly compared to the large decrease in far field true strain at failure, see Figure 4.18. From our experimental observations it can be suggested that pre-strain affects strain localization due to the lower work hardening capability of the matrix and thus plays a major role in driving the fracture process in hole-containing pre-strained material. The only exception is related to the case with pre-strain value of 11.8% where the thickness of the sample was decreased too much in polishing steps and it does not follow the trend line.

4.3.5.3 Digital image correlation map

2D strain maps are based on local derivative calculations which display the strain tensors across the entire tensile surface. We are able to place a virtual strain gauge anywhere on the sample surface after the test, retaining precise strain data. From Figure 4-20 (a) to (d) for different tensile samples, one can see that the pre-strain value changes randomly along the y-axis while the whole gauge was pulled in the same conditions. This strain variation corresponds to shear band formation which propagates randomly along the y-axis with 45° angle with respect to the tensile direction, see Figure 4.18 (c) to (f).

Moreover, Figure 4-23 shows that local strain between voids decreases by increasing the pre-hardening in the material. This can be due to the fact that the sample with higher pre-strain history is more hardened and can reach the failure point more rapidly.

As expected, the global failure strain extracted from the tensile machine shows a lower value compared to the local strain between the voids (taken from DIC maps). Figure 4-26 shows the comparison between the far field true strain and local strain for different pre-strain histories. This significant reduction in far field true strain can be explained in terms of strain localization between the voids.

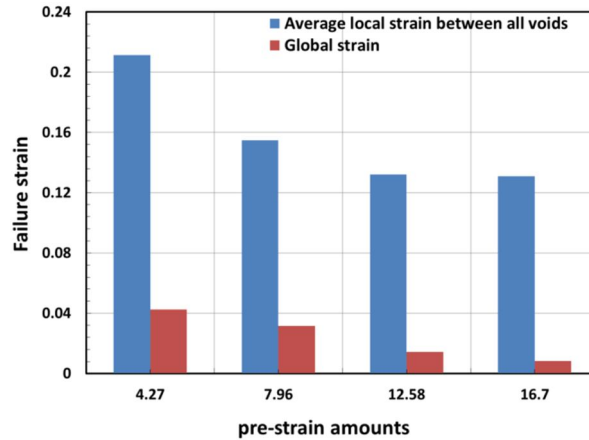


Figure 4-26: Plot of strain comparison between local strain between the voids (extracted from DIC technique) and Global strain (recorded by tensile machine)

4.3.5.4 Major void diameter growth and coalescence (after change in strain path)

Void growth and coalescence were studied for aluminum alloy 5052 under strain path changes in terms of normalized major void diameter (a/a_0) evolution versus far field true strain and strain to failure for different pre-strain histories.

The same trend as in the pure pre-strain case (without strain path changes) has been seen where void growth accelerates by increasing the amount of pre-strain. This behavior can be explained again in terms of the work hardening rate in the samples, i.e. at high pre-strain values the work hardening rate is low and voids grow faster. The scattered experimental data in the case of 8% and 16.64% are related to the decrease in the thickness of the tensile sample due to over-polishing and due to the random shear band propagation along the gauge length of the sample.

The void growth increases gradually in a stepwise manner, similar to the pure pre-strain tests. The observation from optical images in the pure pre-strain

tests is also observed here, where random shear band propagation is responsible for this step change behaviour.

Figure 4-25 shows that strain path change has a negative effect on coalescence strains and is even worse at high pre-strains. As it was explained in pure pre-strain test, it can be due to the dislocation accumulation which reduces the dislocation movement and lowers the work hardening rate which in turns promotes strain localization.

4.4 Comparison between pre-strain and strain path changes

4.4.1 Void growth and coalescence strain

The effect of pre-strain and strain path changes on ductile fracture were studied in sections 4.2 and 4.3. It was found that for the pre-strained samples the yield stresses and the strains to failure were significantly decreased by changing the reloading step 90 degrees with respect to the pre-strain direction. To better compare pure pre-strain and strain path change tests, the strain to failure is plotted versus the amount of pre-strain in Figure 4-27.

A linear relation between strain to failure and pre-strain amount can be seen for the pure pre-strain tests, while the strain path changes show a hyperbolic tensile behaviour, see Figure 4-27. This hyperbolic reduction in strain to failure amount has been explained in terms of work hardening rate by D.V. Wilson et al. (89) (77). At high pre-strains the reorganization of dislocation cells in the new direction is difficult and the prominent transient reduction in work hardening rate happens. Beyond this point, the existence of residual stress from pre-strain stage results in recovering the work hardening rate. However, this increase in work

hardening rate after significant transient reduction at high pre-strain value is relatively small compared to the work hardening rate at low pre-strain value.

Finally, as shown in Figure 4-27, it can be concluded that the aluminum alloy 5052 tensile coupon with 7% pre-strain (half of the normal failure point) shows the largest difference in failure strain between the pure pre-strain and the strain path change tests.

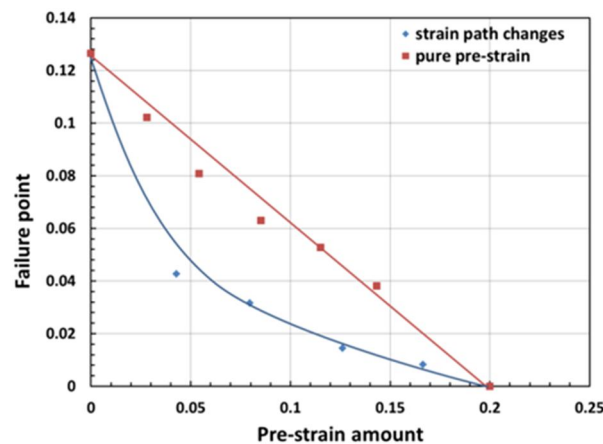


Figure 4-27: Plot of strain to failure versus pre-strain amount; comparison between pure pre-strain and strain path changes for aluminum alloy 5052

4.4.2 True fracture strain

From the SEM analysis it was found that for both pure and strain path changes tests, strain localization plays a major role in driving the fracture processes. Figure 4-28 shows the comparison between pure- pre stain and strain path changes tests where the strain path changes value for true fracture strain has higher value in compare to pure one. This difference in ductility is due to the fact that in strain path changes test, the isotropy sources cause to lower hardening and the material tends to fail earlier in compare to pure one. So, the strain localized more and the sample

fails quickly and the fracture surface necks down to the point and final fracture surface is smaller in compare to the pure-one and consequently the true fracture strain becomes larger due to the equation 4-1.

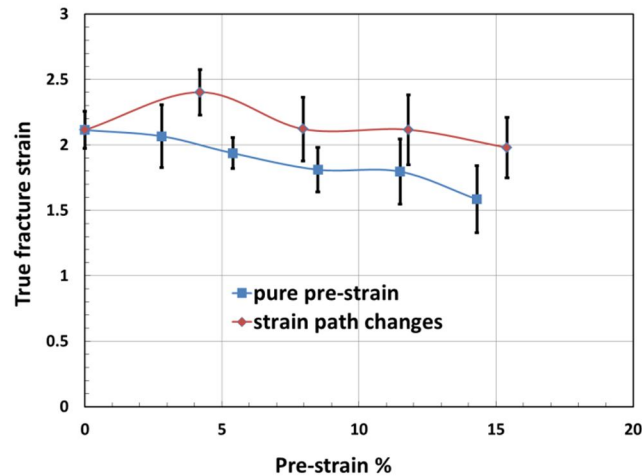


Figure 4-28 Plot of true fracture strain versus pre-strain amount; comparison between pure pre-strain and strain path changes for aluminum alloy 5052

4.5 Conclusion

The main conclusions of this chapter are summarized below:

- Both yield stress and ultimate tensile strength increased by increasing the amount of pre-strain. We concluded that since the dislocation density increases with plastic deformation, the material will experience more hardening and a higher yield stress is needed to resume the plastic deformation (pure pre-strain test).
- SEM images revealed that the ductility of the sample decreased with increasing pre-strain but only slightly compared to the large decrease in far field strain at failure. This suggests that pre-strain affects strain localization significantly and, to a lesser extent, the ductility (pure pre-strain test).

- Void growth was investigated for various amounts of pre-strain in aluminum alloy 5052 (for both pure pre-strain and strain path change). It was shown that increasing pre-strain results in faster void growth which was justified in terms of a decrease in work hardening rate in the sample.
- The gradual increase in void growth in a stepwise manner is due to random shear band propagation through the gauge length of the void-containing pre-strained sample (for both pure pre-strain and strain path changing).
- Void coalescence strain is reduced significantly by increasing the pre-strain history, where dislocation accumulation results in reducing the work hardening rate and increasing the propensity for strain localization.
- Experimental results show a significant reduction in yield stress after strain path changes and a gradual yielding. The main reasons for this behavior are plastic anisotropy, cell block boundaries polarization, strain incompatibilities between grains and the presence of local back stresses in the reloading part are potential reasons proposed in the literature.
- Local strain between the voids is decreased by increasing the pre-hardening in the material, since the sample with higher pre-strain history is more hardened and can reach the failure point more rapidly.
- Digital image correlation maps revealed that local strains between voids have much higher values compared to the far field true strain. This difference is due to the strain concentration between the voids.
- From comparison between pre-strain and strain path changes tests, it was demonstrated that aluminum alloy 5052 behaves differently depending on

the loading condition. With a pre-strain of 7%, the maximum difference in failure strain is observed between pure pre-strain and orthogonal strain path change.

- The true fracture strain for strain path changes has higher value in compare to pure one. This difference in ductility is due to the fact that in strain path changes test, the isotropy sources cause to lower hardening and the material tends to fail earlier in compare to pure one.

Chapter 5

Finite Element Modeling

5.1 Introduction

Finite element analysis has been applied in order to see whether the previous damage modelling methods are able to predict damage by void growth and coalescence. The finite element software Abaqus v6.5 is utilized. Initially we used the previous studies on AA5052 which were using a nonlocal damage model method with a Rice and Tracey type damage indicator. We then found that the material properties were different than those used in previous studied on AA5052. We therefore used the same technique with our experimental material properties using new Voce law fitting parameters. The disagreement between experimental data and simulation results encouraged us to change the fitting equation to the Hollomon law, which conformed well to our experimental data.

Finally, the finite element results using the Hollomon law without damage model was found to be sufficient to predict ductile fracture and are presented and compared with the experimental results at the end of this chapter.

5.2 Modeling of ductile fracture

According to previous work on modeling of aluminum alloy 5052 (81) where a nonlocal damage technique was used to model the void-containing material, the same method was used to simulate the current project which is described and shown in detail in following subsections.

5.2.1 Simulation with damage model approach

The two main approaches to model ductile fracture are surface-based models (fracture mechanics field) and volume oriented based models (continuum mechanics field). In the surface-based approaches, crack propagation occurs due to decohesion between continuum elements or along internal discontinuities provided by special elements (2) (3) (4), while in the volume oriented models damage is defined as the loss of stress carrying capacity in three dimensional elements and their removal from the simulation. The loss of stress carrying capacity behaves as an extension of the J_2 -plasticity theory and is defined through the yield function ϕ as follow:

$$\phi = \frac{\sigma_{eq}}{\bar{D}} - \sigma_f \quad \text{Equation 5-1}$$

Where σ_{eq} is the von Mises equivalent stress, σ_f is the flow stress and $\bar{D}$ is a parameter that shows the progress of damage and varies between 0 and 1. When $\bar{D}=1$ the material is safe whereas at $\bar{D} \rightarrow 0$ the ductile failure happens. The usual definition of $\bar{D}$ is:

$$\bar{D} = \begin{cases} 0 & \text{for } D \leq D_{critical} \\ 1 & \text{for } D \geq D_{critical} \end{cases} \quad \text{Equation 5-2}$$

The parameter D is called “ductile damage indicator” and represents the ductile fracture stages.

5.2.1.1 Ductile damage indicator

There are numerous ductile damage indicators in the literature which are based on the attainment of a critical stress, or strain, or stress triaxiality. The damage indicator proposed by Gunawardena et al. (5) is widely used in continuum mechanics field, which is based on the work of Rice and Tracy (6) and Hancock and Mackenzie (7). This ductile damage indicator includes both strain to failure in uniaxial tension (ϵ_0) and stress triaxiality together as follows:

$$D_i = D_{Gu} = \int_0^{\epsilon_{eq}^{pl,f}} \exp\left(\frac{3}{2} \frac{\sigma_m}{\sigma_{eq}}\right) / 1.65 \epsilon_0 d\epsilon_{eq}^{pl} \quad \text{Equation 5-3}$$

Where ϵ_{eq}^{pl} is the equivalent plastic failure strain. Figure 5-1 shows the failure curve based on the Gunawardena ductile damage indicator. One can see that for an element to be deleted from the mesh, a critical strain and stress triaxiality need to be reached.

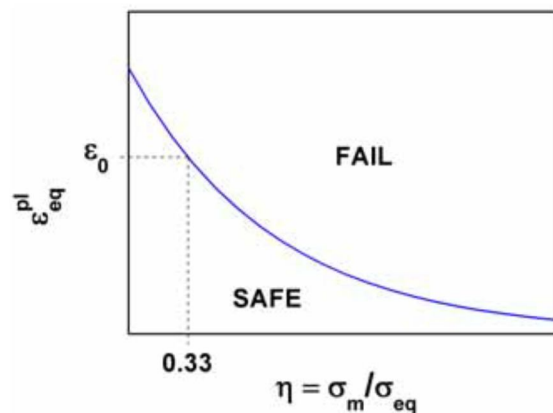


Figure 5-1: Failure curve based on the Gunawardena ductile damage indicator (1).

5.2.1.2 Mesh size dependency

There are several studies in the literature which indicates that finite element results based on strain softening models are significantly mesh sensitive. It was found that strain localizes into a band whose width depends on the element size and tends to zero as the mesh is refined (90) (91). For instance, Drabek and Bohm (91) ran two different damage models based on Gunawardena damage indicator and Gurson-Tvergaard-Needleman. The high mesh sensitivity seen can lead to a large variation in failure strains. This mesh size dependency problem requires one to introduce the “nonlocal damage” technique where the FEM results are not mesh sensitive anymore.

5.2.1.3 Nonlocal regularization technique

Nonlocal regularization techniques act as localization limiters and incorporate a characteristic length (L) into the governing equation. However, choosing the appropriate characteristic length is not simple and can be based on physical values such as size of the process zone, the distance between individual voids, etc. According to previous work on AA5052 (81) the integral method has been used, which is based on the smoothing of the rate of a damage variable $\dot{q}$ over a characteristic length L by nonlocal averaging using a moving weighted-averaging scheme, see Figure 5-2 (a). In other words, for deletion of one element from FEM, the average value of adjacent elements is considered for attainment of critical strain or stress triaxiality. The smoothing function $w(x,y,L)$ which is widely used in the literature is defined in equation 5-4 and is plotted in Figure 5-2 (b).

$$w(x, y, L) = \left[\frac{1}{1 + \left(\frac{|x-y|}{L} \right)^8} \right]^2 \quad \text{Equation 5-4}$$

Where x is the “reference” point and y is the position of the adjacent material points.

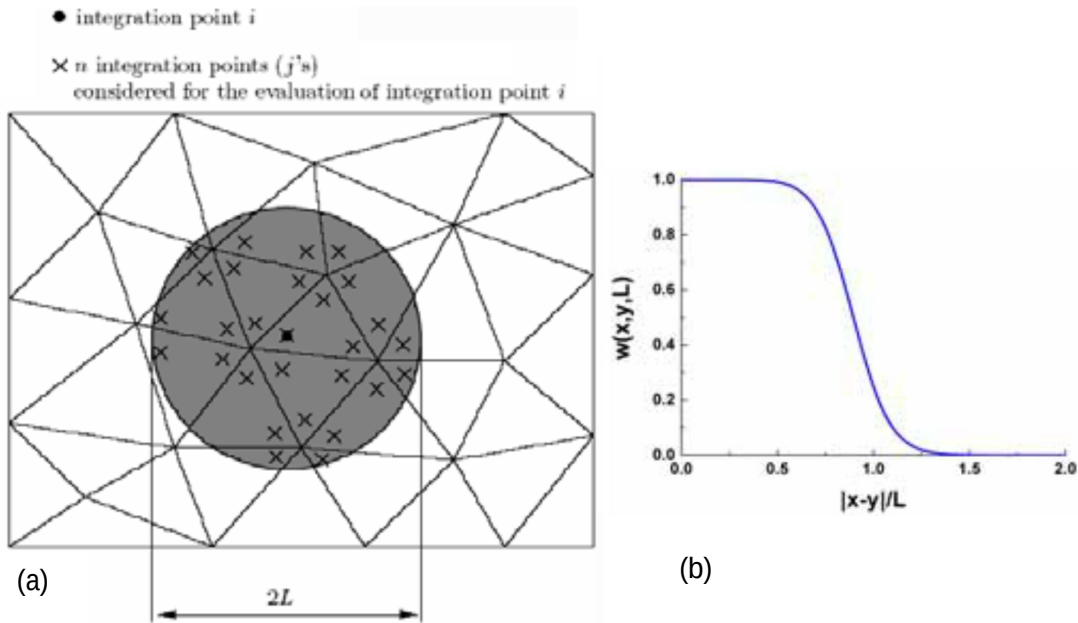


Figure 5-2: (a) Definition of the characteristic length. (from Drabek (2005))(b)Tvergaard and Needelman smoothing function.

The implementation of the nonlocal damage model into the finite element code used in this chapter is not detailed here as it has been previously explained in detail by Drabek and Bohm (91) and Weck et al. (81).

This non-local damage has only two adjustable parameters, namely the local failure strain in uniaxial tension (ε_0) and the characteristic length L which intervenes in the nonlocal averaging scheme in equation 5-4. The parameter ε_0 which is also called Rice and Tracy (RT) parameter controls the point at which damage starts to accumulate in the material and the characteristic length L gives the rate at which damage evolves. However, ε_0 and L are coupled i.e. when the characteristic length

increases, damage is averaged over a larger area and thus the parameter ϵ_0 has to be changed accordingly. This non-local damage model was already used successfully to predict void growth in aluminum alloys 5052 (81).

5.2.2 Comparison between nonlocal damage FEM results with experimental data

In this work the characteristic length L was chosen to be equal to the diameter of the voids ($L=17.5 \mu\text{m}$) and the damage value was set to a strain (RT) of 0.4 as in previous study on aluminum (81). Due to symmetry the analysis was performed on only a quarter of the complete sample geometry used experimentally, see Figure 5-3. Boundary conditions include a Y-symmetry along the xz surface (facing surface in Figure 5-4 that cuts through the holes), X-symmetry along the yz surface, and fixed displacement applied to the upper-end surface equal to 3.75 mm (14 % strain). The simulation has 4949 linear 8-nodes brick fully integrated (C3D8) elements and the mesh was refined around the holes, making sure that the characteristic length contains at least 3 elements (for removing mesh sensitivity). The dimension of the model is 1.5mm x 0.5 mm x 0.103 mm which reflects the real experimental sample dimensions taking into account the symmetry (3mm x 1mm x 0.103).

The previous Voce law parameters used by A. Weck et al. (81) have proven to be successful for aluminum alloy 5052 and were chosen for the current study as the same material was used. The fitted parameters for the Voce law are presented in Table 5-1. Figure 5-4 presents images taken in 4 different time-steps during the simulation and shows how crack starts in the middle of the sample and propagates

until fracture. The grey elements in Figure 5-4 are elements that reached the critical damage value and therefore do not carry anymore load. These elements simulate the ductile fracture process taking place between voids.

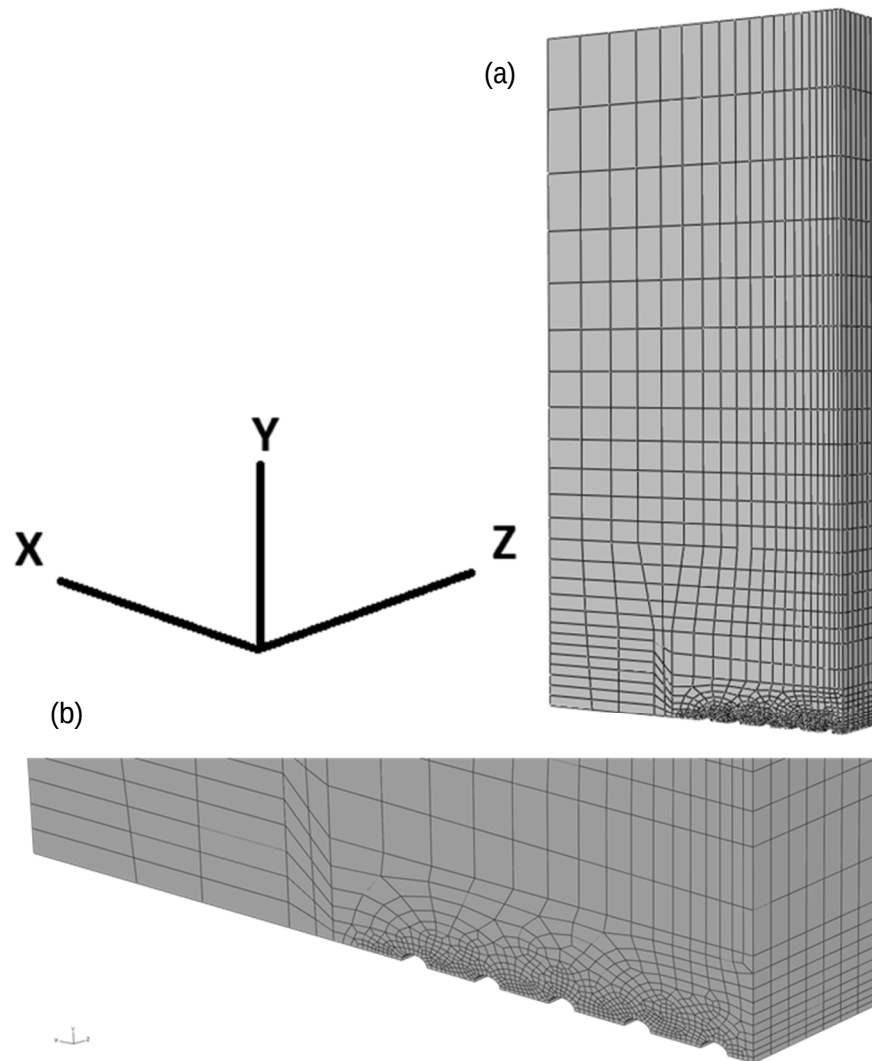


Figure 5-3: FEM 3D-model for 5 holes configuration at 90° with respect to the tensile axis (along direction y). Only 1/4th of real sample is modelled; a) snapshot from whole mesh b) mesh near the holes

Table 5-1 Fitting parameters for the Voce behaviour of aluminum alloy 5052

Law	Equation	parameters
Voce	$\sigma = a - b \exp(-c\varepsilon)$	$a=445 \text{ MPa}, b=300 \text{ MPa}, c=5.4$

The simple nonlocal ductile fracture model presented here is compared with experimental data in Figure 5-5 and Figure 5-6 for aluminum alloy 5052 with different pre-strain histories. On one hand Figure 5-6 (a) to (d) shows that the nonlocal damage model with previous Voce law parameters (81) is able to capture the behaviour observed experimentally in terms of void growth and failure strains for different pre-strain histories. Conversely a large discrepancy between FEM results and experimental data in stress-strain curves can be seen in Figure 5-5 (a) to (d).

Although the current Voce law parameters have been successfully used by Weck et al. (81) for the same material, it appears that new Voce law parameters are required. The strain results are significantly affected by the rigidity factor of the tensile machine used to carry out the test which may not have been taken into account in previous studies. In this study, a rigidity factor of 1.5 was experimentally determined.

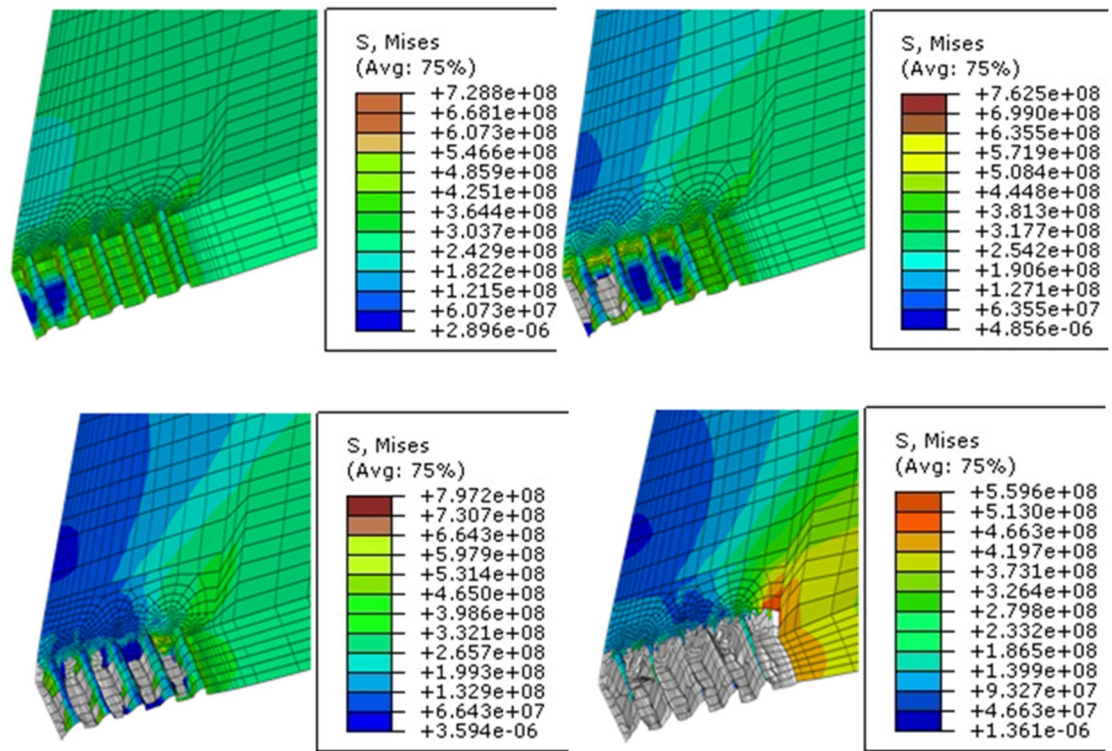


Figure 5-4: Element deletion process in ABAQUS using non-local damage model; grey parts are elements that reached the critical damage value and therefore do not carry any more load

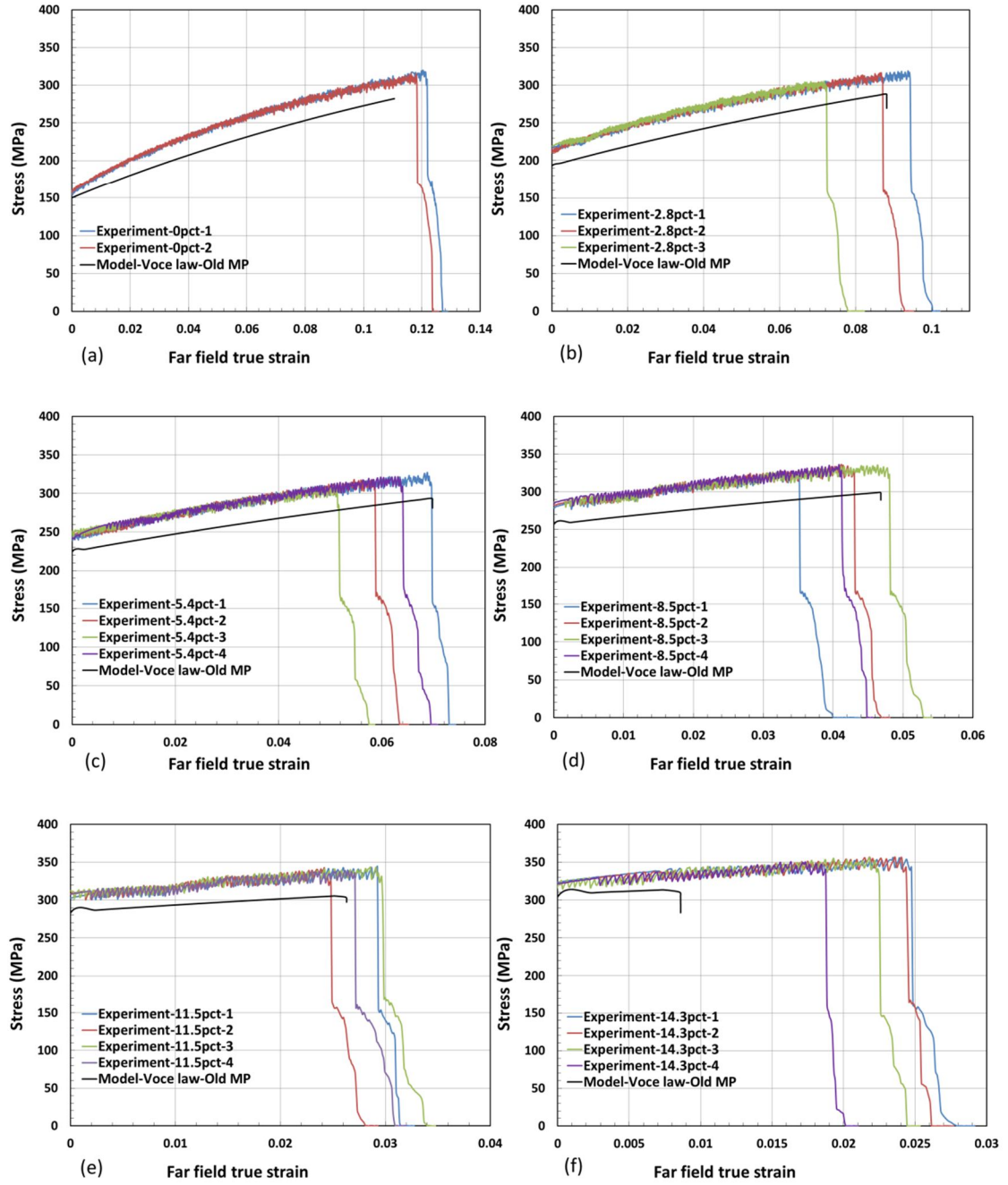


Figure 5-5: Comparison between experimental results and nonlocal damage finite element simulation for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Rice and Tracy damage parameter of 0.4 and characteristic length equal to $17.5 \mu\text{m}$ are chosen for nonlocal technique within old material properties from Voce law

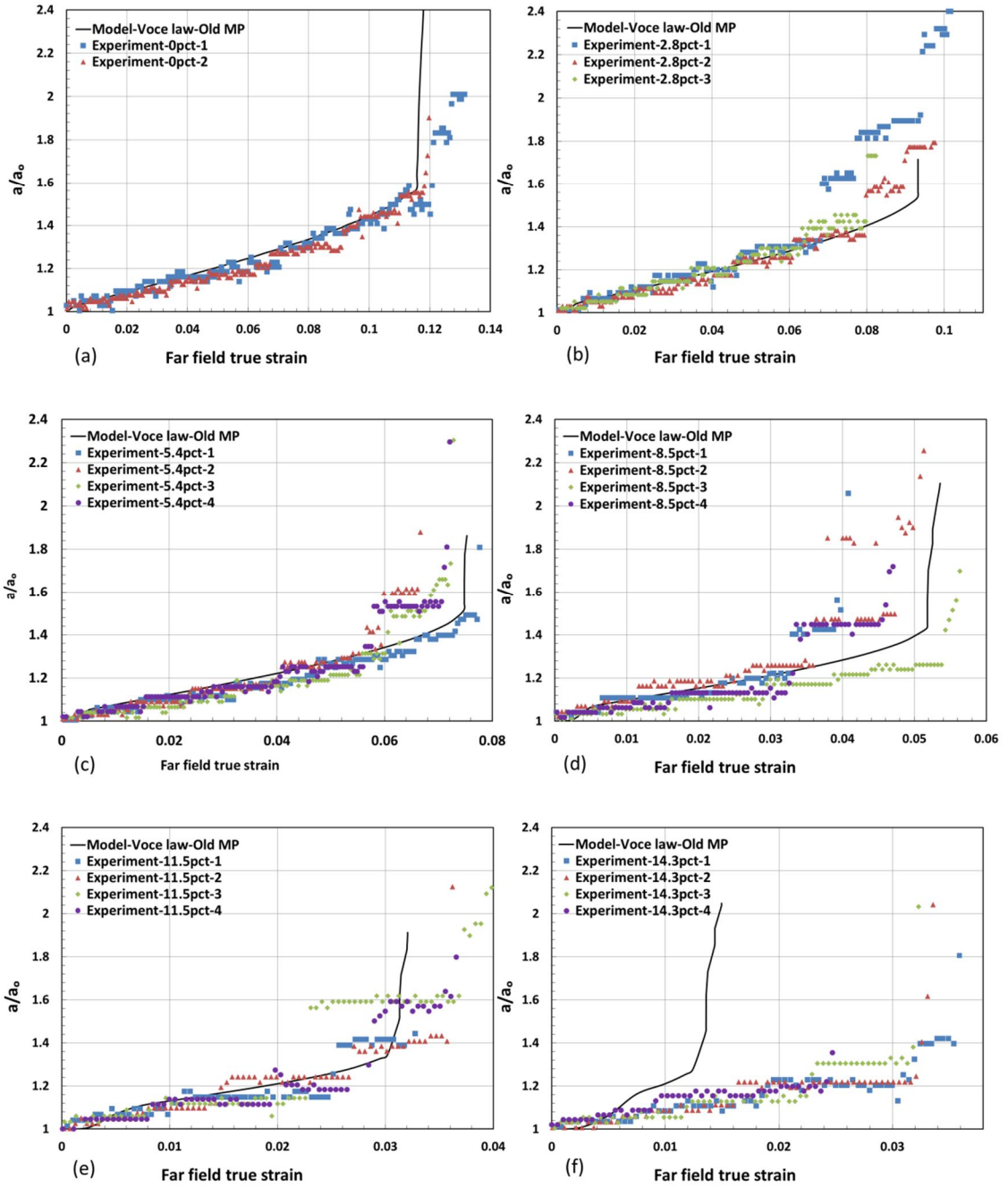


Figure 5-6: Comparison between experimental results and nonlocal damage finite element simulation for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Rice and Tracy damage parameter of 0.4 and characteristic length equal to $17.5 \mu\text{m}$ are chosen for nonlocal technique with old material properties from Voce law

5.2.3 Comparison between nonlocal damage FEM results with experimental data; using new Voce law parameters

The true stress-true strain curve of 0% pre-stained aluminum alloy 5052 tensile sample without hole was used to extract the new Voce law parameters. The Voce equation is fitted to current experimental results in Figure 5-7. The new fitting parameters for the Voce law are presented in Table 5-2.

Table 5-2: New fitting parameters for the Voce behaviour of aluminum alloy 5052

Law	Equation	New parameters
Voce	$\sigma = a - b \exp(-c\epsilon)$	$a=390 \text{ MPa}, b=267 \text{ MPa}, c=9.67$

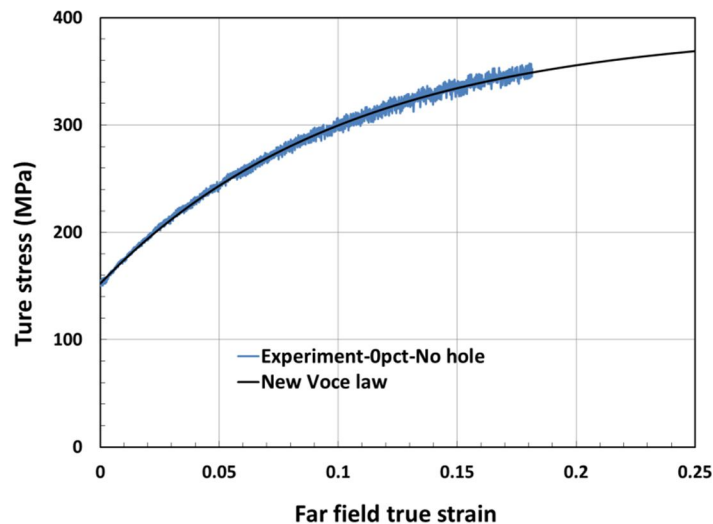


Figure 5-7: Comparison between the new Voce law and the experimental results

Comparison between the nonlocal damage model using the new Voce law fitting parameters and the experimental data is presented in Figure 5-9 and Figure 5-10. One can see that even though the FEM better predicts the stress strain curve, it is not able to predict neither void growth nor failure strain and the prediction is worse than with the previous parameters. In this case the FEM did not capture the

experimental data for void growth, particularly at higher pre-strain amounts which means that earlier coalescence occurred in the finite element model.

Comparison between the two Voce law fitting parameters (old and new) in Figure 5-8 indicates that for the new fitting parameters the hardening goes to zero after the strain value of 0.5, which results in earlier coalescence predictions in Figure 5-10. The work hardening rate for the old Voce law after a strain value of 1 tends to zero which leads to a better prediction of void growth.

Therefore for better damage predictions in term of void growth and stress-strain curve, particularly at higher pre-strain value, more strain hardening is needed in the finite element model. For this purpose, the Hollomon law was used and fitted to our experimental data where the strain hardening does not tend to zero.

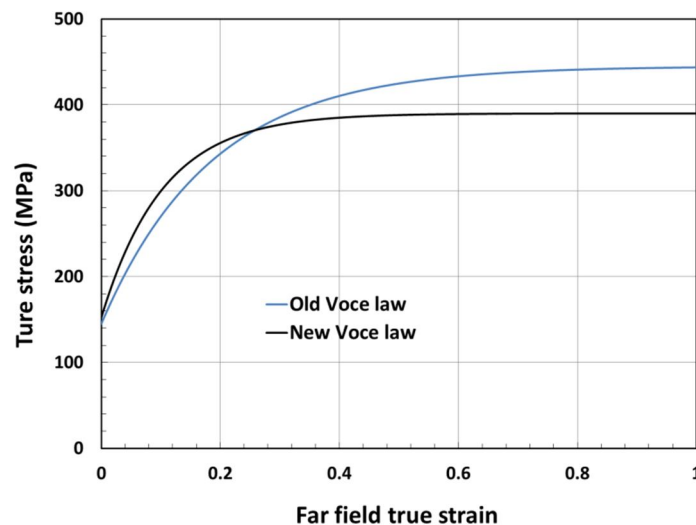


Figure 5-8: Comparison between the new Voce law and the old one

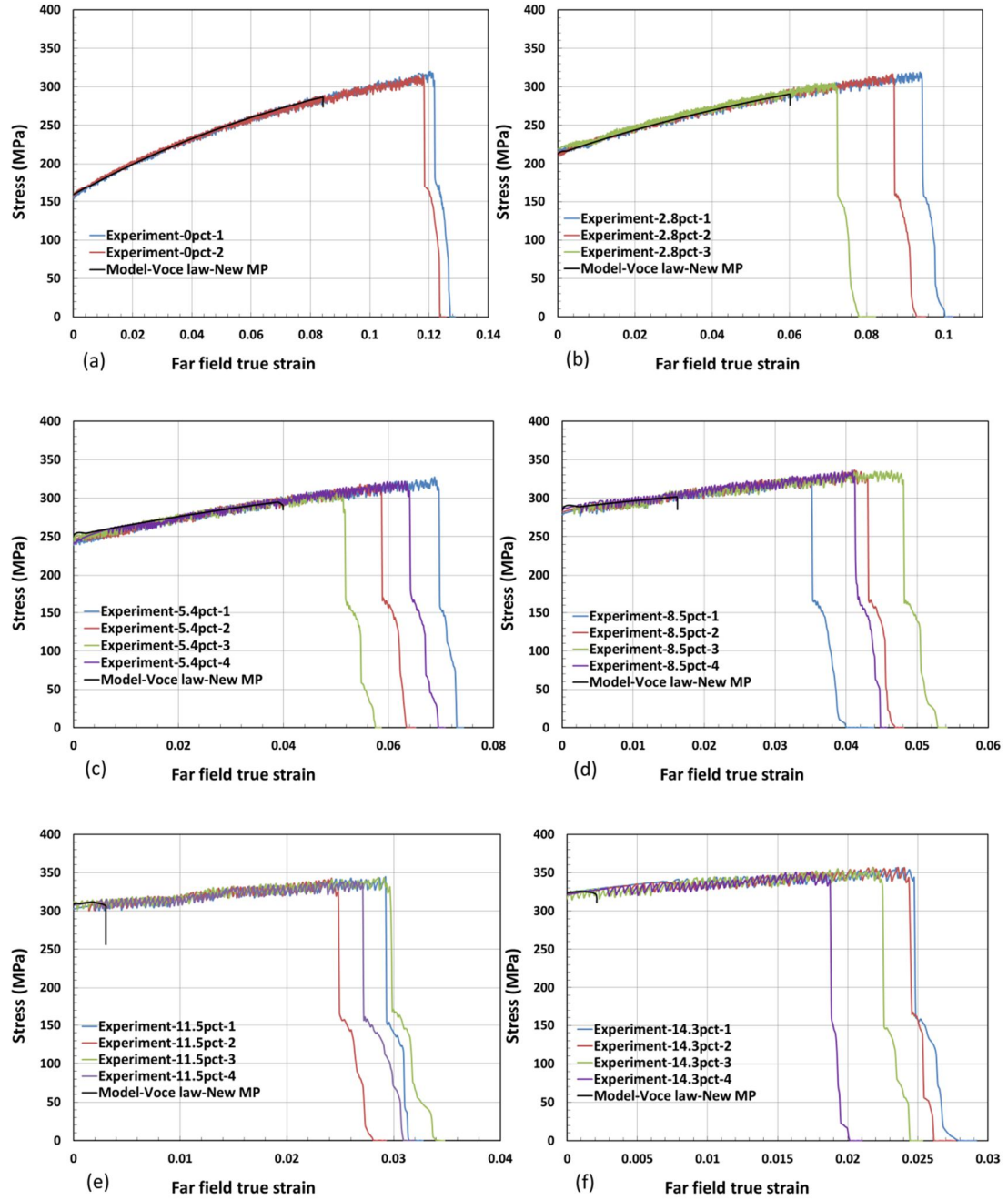


Figure 5-9: Comparison between experimental results and nonlocal damage finite element simulation for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Rice and Tracy damage parameter of 0.4 and characteristic length equal to $17.5 \mu\text{m}$ are chosen for nonlocal technique with new material properties from Voce law

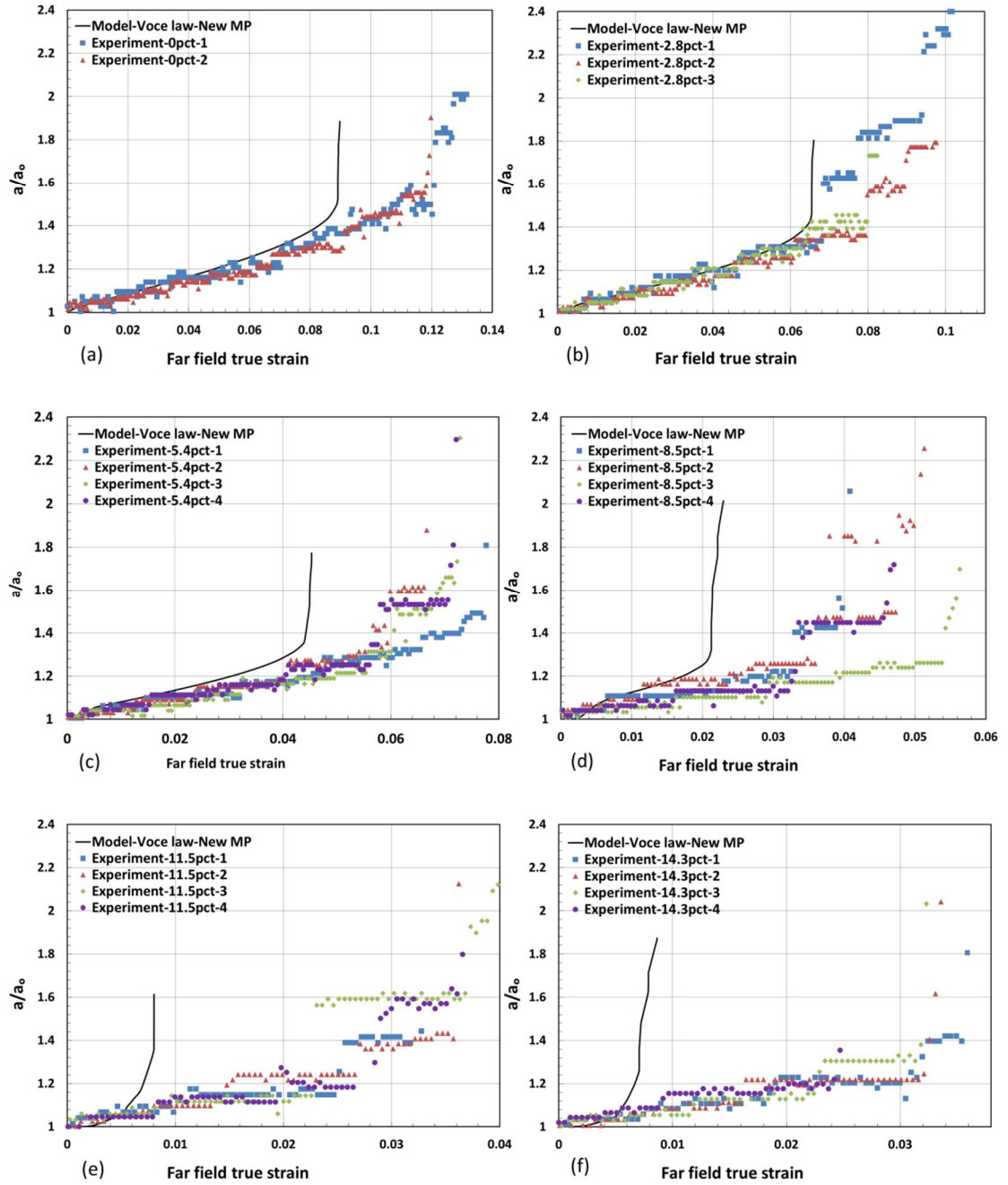


Figure 5-10: Comparison between experimental results and nonlocal damage finite element simulation for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Rice and Tracy damage parameter of 0.4 and characteristic length equal to $17.5 \mu\text{m}$ are chosen for nonlocal technique with new material properties from Voce law

5.2.4 Comparison between nonlocal damage FEM results with experimental data; using Hollomon law fitting parameters

The true stress-true strain curve of non-hardened aluminum alloy 5052 tensile coupon without hole was used to obtain the Hollomon law fitting parameters. The Hollomon equation is fitted to the experimental data in Figure 5-11. The parameters for the Hollomon law are shown in Table 5-3.

Table 5-3: Fitting parameters for the Hollomon behaviour of aluminum alloy 5052

Law	Equation	New parameters
Hollomon	$\sigma = K\varepsilon^n$	K=604.48, n=0.3252

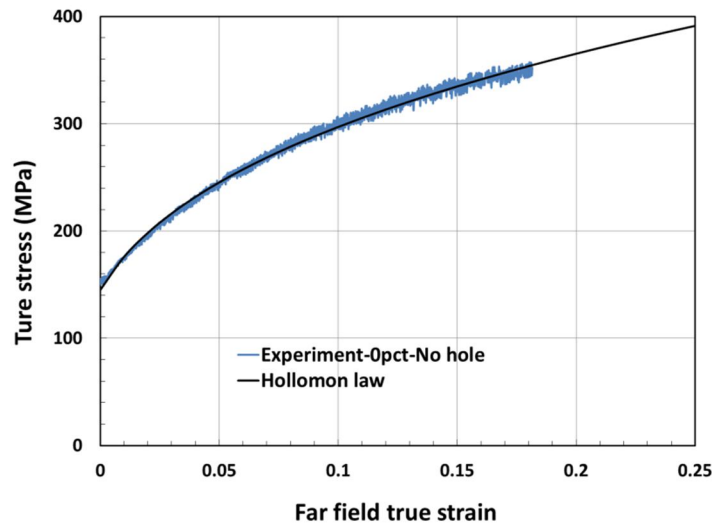


Figure 5-11: Comparison between the Hollomon law and the experimental results

Comparison between the nonlocal damage model using the Hollomon equations and experimental data are presented in Figure 5-12 and Figure 5-13 for the stress-strain curve and void growth respectively. One can see that the simulation predicts well the experimental void growth as well as the coalescence part, even at high pre-strain amount. The only discrepancy is found between the simulation and experimental results for the case of 14.3% pre-strain.

It also was found that the onset of coalescence in the finite element analysis with damage model using the Hollomon law is beyond the strain where we expected it to happen experimentally, see Figure 5-13. From this point of view, one can conclude that the FEM using Hollomon fitting parameters does not need the nonlocal damage model to predict the void growth and coalescence in aluminum alloy 5052. Hence the FEM analysis without damage model was used and compared with experimental data in Figure 5-14 and Figure 5-15. The accurate prediction confirmed the latter arguments where it is said that the damage model is not needed to model void growth and coalescence in aluminum alloy 5052.

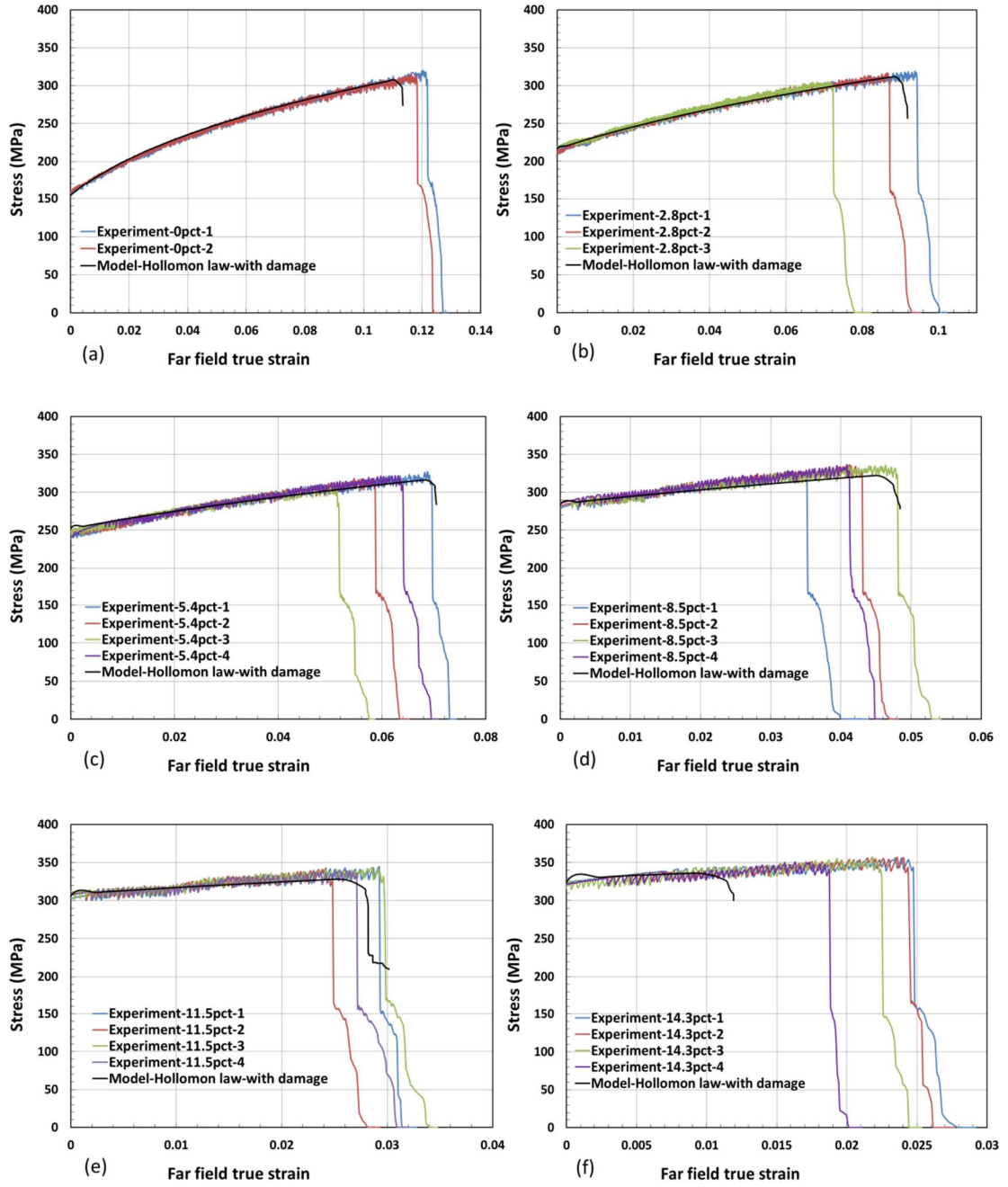


Figure 5-12: Comparison between experimental results and nonlocal damage finite element simulation for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Rice and Tracy damage parameter of 1 and characteristic length equal to $17.5 \mu\text{m}$ are chosen for nonlocal technique with new material properties from Hollomon law

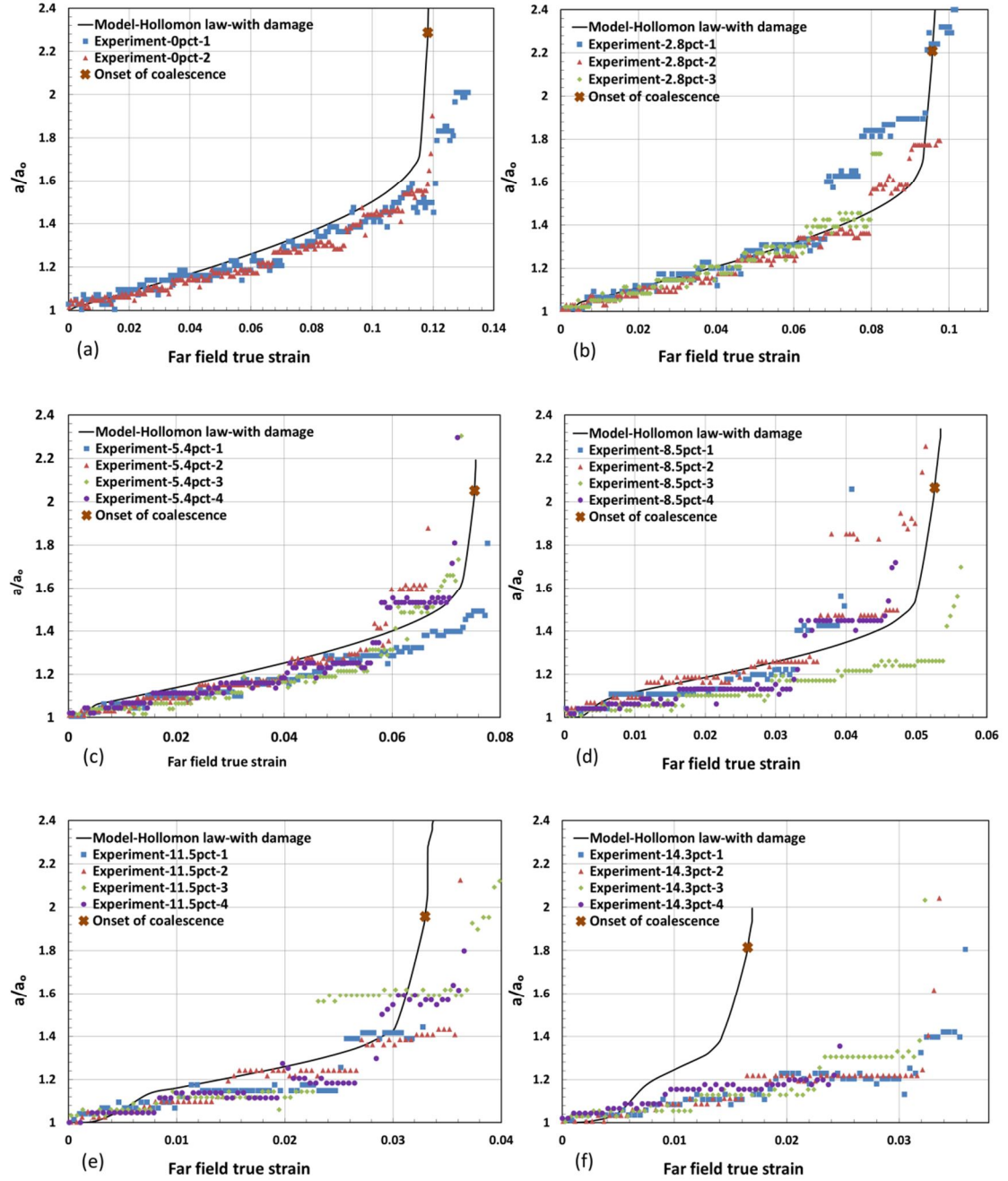


Figure 5-13: Comparison between experimental results and nonlocal damage finite element simulation for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Rice and Tracy damage parameter of 1 and characteristic length equal to $17.5 \mu\text{m}$ are chosen for nonlocal technique with new material properties from Hollomon law. The Celtic cross point indicate the onset of coalescence

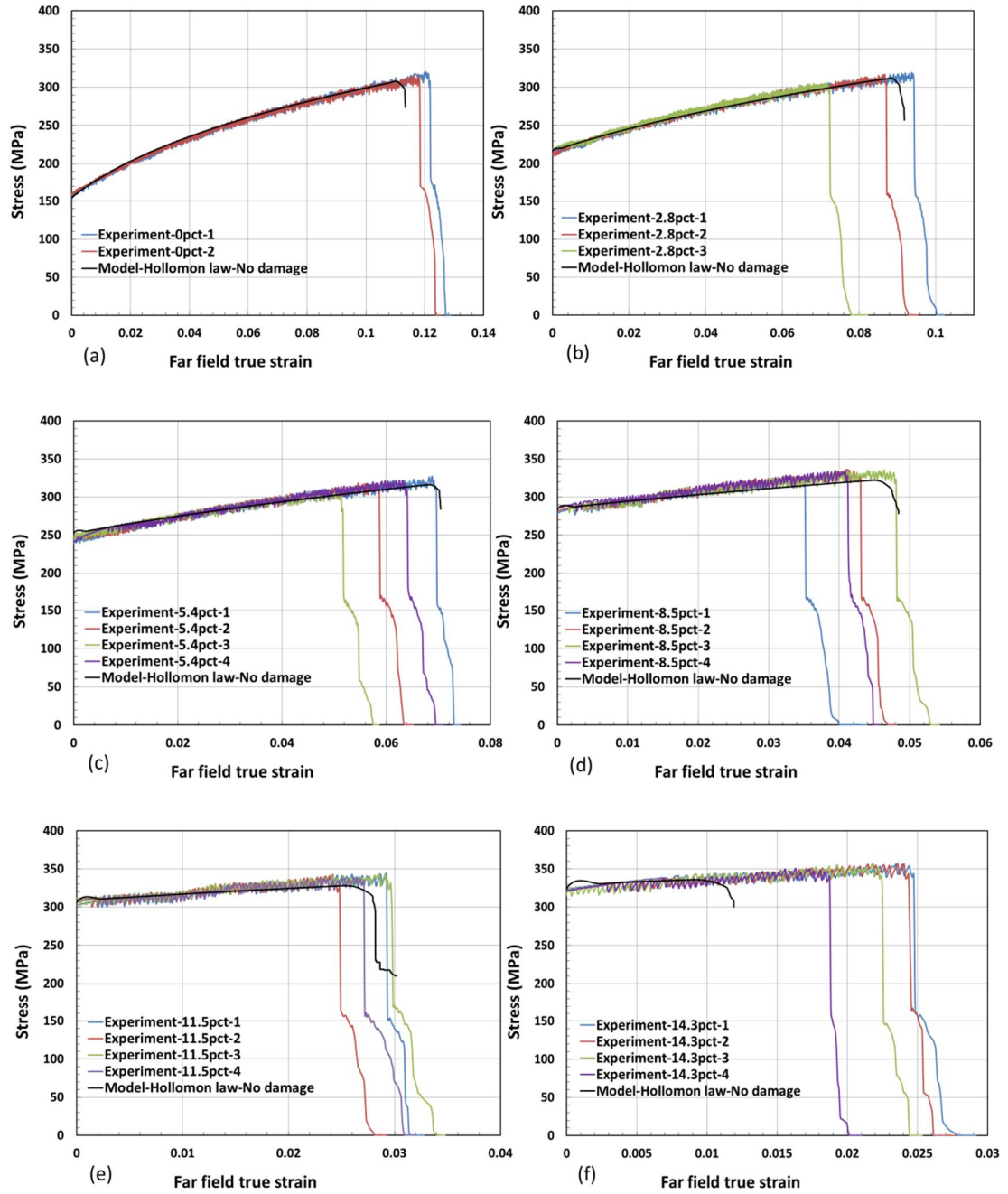


Figure 5-14: Comparison between experimental results and simple finite element simulation without damage for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Hollomon law constants were used for plastic material properties.

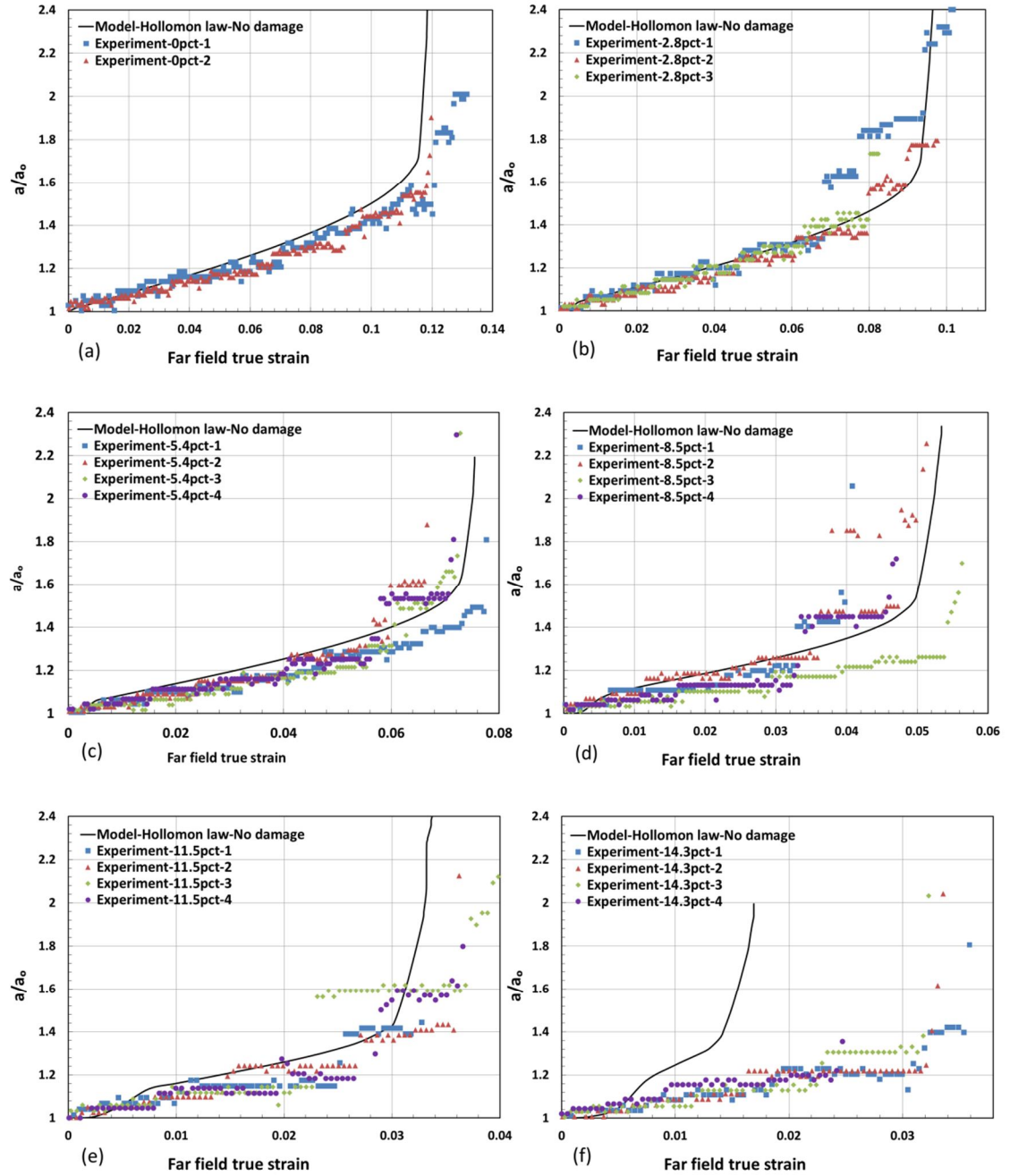


Figure 5-15: Comparison between experimental results and simple finite element simulation without damage for different pre-strain histories; a) 0%, b) 2.8%, c) 5.4%, d) 8.5%, e) 11.5% and f) 14.3%. Hollomon law constants were used for plastic material properties.

5.2.5 Conclusion

The conclusions for this chapter are listed below:

- The 3D finite element model was utilized for predicting void growth and coalescence in aluminum alloy 5052
- To model damage we concentrated on the element deletion technique using a Rice and Tracey type damage indicator and a nonlocal damage formulation. This non-local damage has only two adjustable parameters, namely the local failure strain in uniaxial tension and the characteristic length L . In this work L was chosen to be equal to the diameter of the voids ($L=17.5\text{ }\mu\text{m}$) and the damage value to 0.4.
- The nonlocal damage model with old Voce law fitting parameters captured the void growth and coalescence well but it was not able to capture the behaviour of stress-strain curves.
- The new Voce law fitting parameters resulted in a worse prediction of void growth and in a better prediction of the tensile behaviour.
- The damage model using Hollomon law fitting parameters with higher strain hardening capacity gives good predictions for both void growth and stress-strain curves.
- We finally showed that it is not necessary to use the damage model when the proper material constitutive equation is used, in this case the Hollomon law.

Chapter 6

Conclusion and future work

The main goal of this thesis was to experimentally visualize in detail void growth and coalescence and assess the accuracy of the McClintock (2D) void growth model and 3D finite element simulations. The main contributions are described below:

- Two main techniques have been utilized in the experimental setups; laser model material and digital image correlation.

Due to the complicated microstructure of commercially available material, model materials using micron-size laser drilled holes were employed to better understand the effect of pre-strain and strain path changes on ductile fracture in aluminum alloy 5052. Additionally the same initial state of void size and void shape are required for better comparison between different pre-strain and strain path change histories.

In order to obtain the real amount of pre-strain and local strain between the voids, digital image correlation (DIC) technique was used where Davis 8.1 software tracks the position of physical points to produce strain maps of the sample surface.

- Behaviour of stress-strain curve

In the case of pure pre-strain (without change in strain path), both yield stress and ultimate tensile strength increased with the amount of pre-strain. We demonstrated that since the dislocation density increases with plastic deformation, the material will experience more hardening and a higher yield stress is needed to resume the plastic deformation.

However experimental result shows a significant reduction in yield stress after the strain path changes due to a softening effect. The main reasons for this softening is plastic anisotropy, cell block boundaries polarization, strain incompatibilities between grains and the presence of local back stresses during reloading.

- Ductility of the material for both pure and strain path changes test

SEM images revealed that the ductility of the sample decreased with increasing pre-strain but only slightly compared to the large decrease in far field strain at failure. This suggests that pre-strain affects strain localization significantly and, to a lesser extent, the ductility.

- Digital image correlation maps for 2D strain field

Digital image correlation maps revealed that local strains between voids have a much higher value compared to far field true strain. This difference is due to strain concentration between the voids.

Local strain between the voids is decreased by increasing the amount of pre-hardening in the material, since the sample with higher pre-strain history is more hardened and can reach the failure point more rapidly.

- Major void diameter growth

It was shown that increasing pre-strain results in faster void growth which was justified in terms of a decrease in work hardening rate in the sample (for both pure pre-strain and strain path changes). The gradual increase in void growth followed a stepwise behaviour which is due to random shear band propagation through the gauge length of the void-containing pre-strained sample.

- Void coalescence and failure strain

The work presented here considered coalescence as the strain at which the ligament between voids is broken. However coalescence is generally characterised as a strong localization between voids and is rather difficult to observe experimentally. Coalescence is defined here as the failure strain of the material.

Void coalescence strain (for both pure pre-strain and strain path changed samples) is reduced significantly by increasing the pre-strain history, due to the lower work hardening rate of samples with higher pre-strain histories.

- Pure pre-strain result versus strain path changes results

Comparison between pre-strain and strain path changes tests indicated that the aluminum alloy 5052 tensile sample with 7% pre-strain shows the maximum difference in failure strain between pure pre-strain and strain path change.

- Finite element modeling

To predict void growth and coalescence using the finite element method, a nonlocal averaging technique was used as a starting point. By using new fitting parameters (Hollomon law), it was found that we do not need the damage model for predicting void growth and coalescence in aluminum alloy 5052. These results show the importance of choosing the proper constitutive equation in finite element models for accurate failure predictions.

Although the results presented here have demonstrated the effect of pre-strain and strain path changes on ductile fracture both experimentally and using finite element models, it could be further developed in a number of ways:

- Development of digital image correlation method

Quality of the 2D strain maps are controlled by two parameters; resolution of the fine features deposited between the voids and their density. In the electrodeposition method, the time and voltage control the size and density of patterns respectively. By using the power supply with higher voltage

capacity ($V > 40\text{V}$) and a transistor for controlling the time (less than 1 sec) it could be possible to get finer features with higher densities.

- 3D experimental results

The 2D experimental results in term of void growth in this study for the laser drilled material could be performed in 3D to see whether the results differ in 3D.

- Effect of crystal structure

The effect of pre-strain and strain path changes on ductile fracture can also be studied for different crystal structures rather than FCC in this work. Due to the different atomic structure in BCC or HCP, the fracture properties of these materials may demonstrate different behaviour.

- Implementation of models into FEM code for strain path changes test

Development of FEM codes including crystal plasticity or dislocation dynamics could provide the necessary tools for predicting strain path change effect.

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