Applying Digital Image Correlation Methods to SAMP-1 Characterization

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1 Abstract

SAMP-1 is a complex material model designed to capture non-Mises yield and localization behavior in plastics. To perform well, it is highly dependent on accurate post-yield material data. A number of assumptions and approximations are currently used to translate measured stress-strain data into the material parameters related to these inputs. In this paper, we look at the use of direct localized strain measurements using digital image correlation (DIC) as a way to more directly extract the required data needed for SAMP-1

2 Introduction

Advanced material models require the supply of information to the simulation that goes beyond the stress-strain data that are usually available. These data provide a more detailed quantification of non-linear behaviours that give the simulation the ability to provide more accurate predictions of post-yield plasticity and eventually failure. The SAMP-1 model [1] is one such model which contains these advanced features. It is not a failure model in that it does not quantify the mode of failure of a plastic; however it is able to provide more accurate quantification of plasticity with an attempt to account for deviatoric and volumetric flow as well as non-Mises behaviour. The quantification of plasticity follows a novel approach where an attempt is made to evaluate the true stress and strain even in the plastic (hardening) region by a series of calculations that use a ‘Poisson’s Ratio’ measured in the plasticity region. By having a measured rather than theoretically estimated cross-section, the stress is more accurate. Rather than assuming the yield locus to be the von Mises envelope, two or more of these datapoints are measured to provide a better estimation of plasticity in shear, biaxial and compressive modes where, with traditional elastic-plastic material models, just the uniaxial tension data would be used.

Digital image correlation is a relatively new technique where it is possible to measure strains in multiple axes simultaneously during a single measurement. We performed detailed studies of the technique to evaluate its ability to capture stress-strain data. Our objective was to see what difference, if any, was obtained using this method as compared to conventional contact methods. This was an important step in order to ‘calibrate’ the technique and show traceability of this new method to the large body of currently existing data that is based on contact extensometry. The next step was to perform experiments that could not be done using contact methods. These pertained particularly to the post-yield Poisson’s Ratio measurements which are at best cumbersome, but also potentially erroneous as we will discuss later. We used DIC to measure multi-axial strains, directly arriving at the desired Poisson’s Ratio measurements without calculations or assumptions. This data simplified the calculations and brought a higher quality of volumetric data to the material model.

The von Mises theory is invaluable in mechanical design because it allows the computation of yield in other deformation modes such as shear and biaxial based on the data from a uniaxial tensile experiment. It is of enormous benefit when modelling materials that follow this behaviour because complexity of the analysis is greatly reduced and the material data required is quite simple. Important exceptions unfortunately exist where the material does not follow this behaviour and in the case of ductile plastics, SAMP model attempts to capture physical behaviours not conforming to the von Mises theory. The experiments are difficult in nature. While a compressive experiment may be manageable, an accurate biaxial tension experiment is extremely difficult in plastics. Some shear experiments have been reported to show tensile characteristics at moderate to large strains. Clearly, the purpose of the data would be defeated should the specimen not be in a state of pure shear at yield. In this case the shear yield stress would be incorrect, making the experiment useless. We used DIC to investigate the Iosipescu shear experiment [2] to see whether a higher pure-shear limit could be achieved.

Finally recent communications [3] indicate an interest in the stacked compression measurement as a possible means to obtain biaxial data because it allows for the ability to attain large stable biaxial strains without localization. We performed these measurements on the polycarbonate.

3 Experimental Work

A polycarbonate material #85805K26 available in 3.2 mm (1/8") thick sheets from McMaster Carr Company [4] was used in the studies. Details of whether the sheet was extruded or cast were not available. Polycarbonate was chosen because it is an amorphous ductile polymer and does not have the additional complexity that arises with semi-crystalline plastics, which exhibit post-yield stress-whitening during the necking phase. A polycarbonate was also used in the original work on SAMP [1]. Test specimens of the desired shapes were cut from the sheet using CNC methods. The data requirements for SAMP include rate dependent tensile stress-strain data, described elsewhere [], plus the following properties measured at quasi-static strain rates:

- Tensile mode Poisson's ratio in the plasticity region
- Shear stress-strain
- Compressive stress-strain
- Biaxial stress-strain

For the quantification of the Poisson's Ratio, ASTM D638 Type I [5] tensile bars were used; the specimens have a 50 mm long gauge length and are 12.5 mm (1/2") wide. Tests were performed using an Instron 8872 servohydraulic universal testing machine (UTM) equipped with 25 KN load cell. A crosshead speed of 5 mm/min was used. Prior to the measurement, each specimen was coated with a speckle pattern to prepare it for DIC. Care was taken to ensure that the coating process did not cause swelling or crazing in the polymer which would result in premature failure. An ARAMIS DIC

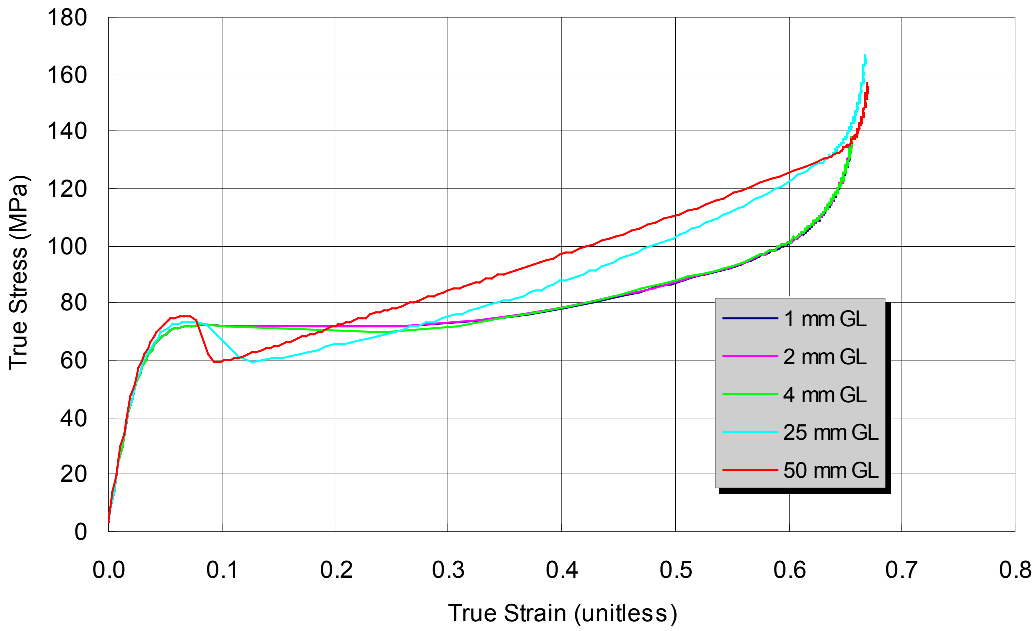

Fig. 1: Selecting a gauge length for the measurement

system was used for the measurements. The apparatus utilizes a stereo video camera to capture three dimensional displacements of the test specimen. For the purposes of this experiment, a 50 mm long, 12 mm wide area of the test specimen was selected within the gauge region. The average strain in the 50 mm direction was used to quantify the axial or longitudinal strain, arbitrarily called the y-strain. The average strain in the 12 mm direction was used to quantify the transverse x-strain. The z-strain in the thickness direction was also measured, where the instrument was able to measure the displacement of the face of the test specimen. An axi-symmetric assumption was used to then calculate the total z displacement. With these measurements, the y, x and z strains were known throughout the measurement rendering it possible to measure the Poisson's ratio at any strain.

We found that a stress-strain curve measured in this way was identical to that measured using contact extensometry. It was observed however, that there was potentially an error in the post yield calculation. As the neck develops, the straining begins to localize so that the idea of averaging strains over the entire 50 mm gage region could be potentially erroneous. To check this, for the same experiment, we recalculated the stress-strain response but this time for a 25 mm gage length. In all cases, care was taken to ensure that the localization occurred within the gauge length. Finally, a 2 mm gage length was used by locating an area within the region of localization and then back-calculating the stress strain response for that region. This resulted in a surprising result shown in Figure 1, where the hump that is typically observed at yield disappeared and a stress-strain curve was obtained that was identical with theoretical predictions. We then perturbed the gauge length, doubling it and halving it, to see if a definitive curve could be obtained. It was found that in this particular case, a 4 mm gauge length was found to reintroduce the hump. The 1 mm gauge length gave virtually the same result, suggesting that it is possible to asymptotically approach the true stress-strain behaviour of a plastic in the post-yield region. Considering that the greatest accuracy would be obtained with the largest gauge length that did not produce a hump, a 2 mm gauge was used for subsequent calculations with this material. The x and z strains were recorded over the entire measurement as shown in Figure 2.

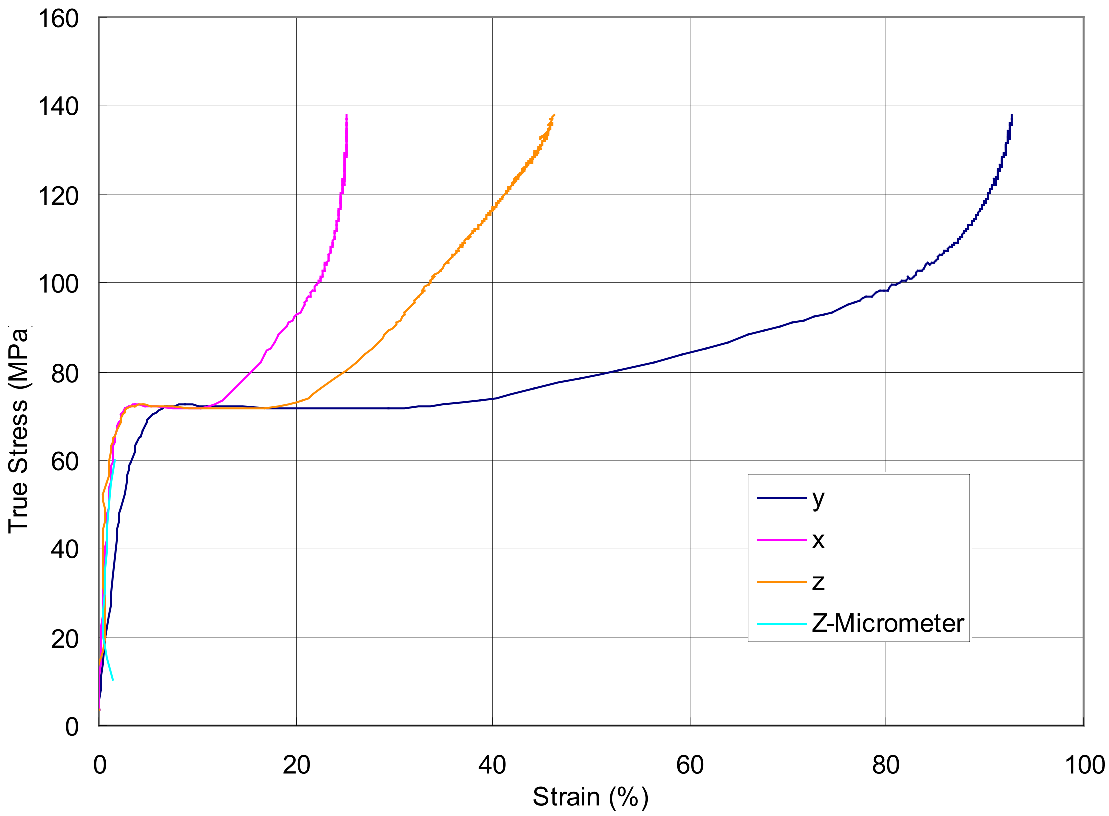

Fig. 2: Strains in the axial (y) and width (y) and thickness (z) directions

We noted that the strains in the transverse x and z directions were identical until yield but then began to diverge from each other post yield. In other words, prior to the onset of localization, the specimen experiences uniform reduction in cross-sectional area with the Poisson's ratio providing a correct translation of longitudinal strain to transverse strain. This is not the case post-yield. We were concerned about the accuracy of the z-direction measurement where displacements are very small coupled with the extreme difficulty of making the out of plane measurement with DIC. In a separate experiment, a NIST traceable micrometer was used to manually measure the thickness of the specimen while subjecting the specimen to finite stresses up to 60 MPa. The z-strains calculated from these manual measurements is seen to be identical with those measured by DIC confirming that our DIC measurements were of adequate precision even in the z-direction.

In an effort to understand the nature of the strain changes at yield, we plotted the strains v. time (Figure 3). We observed a rapid change in longitudinal local (y) strain upon the onset of yield that substantiated the theory proposed by Diehl [6].

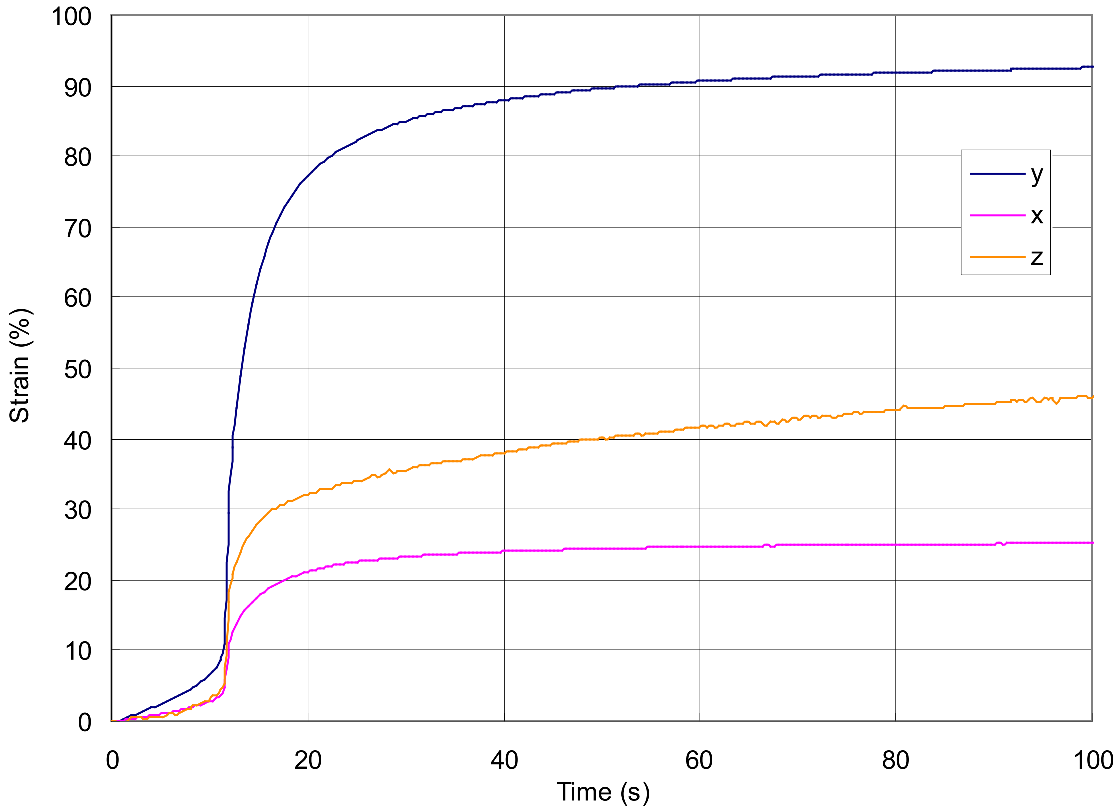

Fig. 3: A rapid increase in strain is noted at yield

Knowing the strains in each direction, it was then possible to calculate a volumetric stress-strain curve as shown in Figure 4. The curve is superposed along with the true stress-strain curve to make it easier to visualize the events as they occurred. We observed that there was zero or possibly compressive volumetric strain in the specimen prior to yield indicating that the material was hydrostatically neutral or possibly in a state of constant hydrostatic compression. In the absence of adequate data, however, no definite statement could be made. Upon the onset of yield, the volume was observed to jump sharply by about 3.5%, then stabilize until a stress of 90 MPa was reached, corresponding to about 55% true strain, after which the volume increased again until failure occurred. It was unclear whether the sharpness of the transition in volumetric strain at the onset of yield was properly captured because the strain evolution at this point is very rapid as can be seen in Figure 2 and there were not enough datapoints to adequately characterize this portion of the curve. It was interesting to note that the volume strain remained constant during the neck propagation.

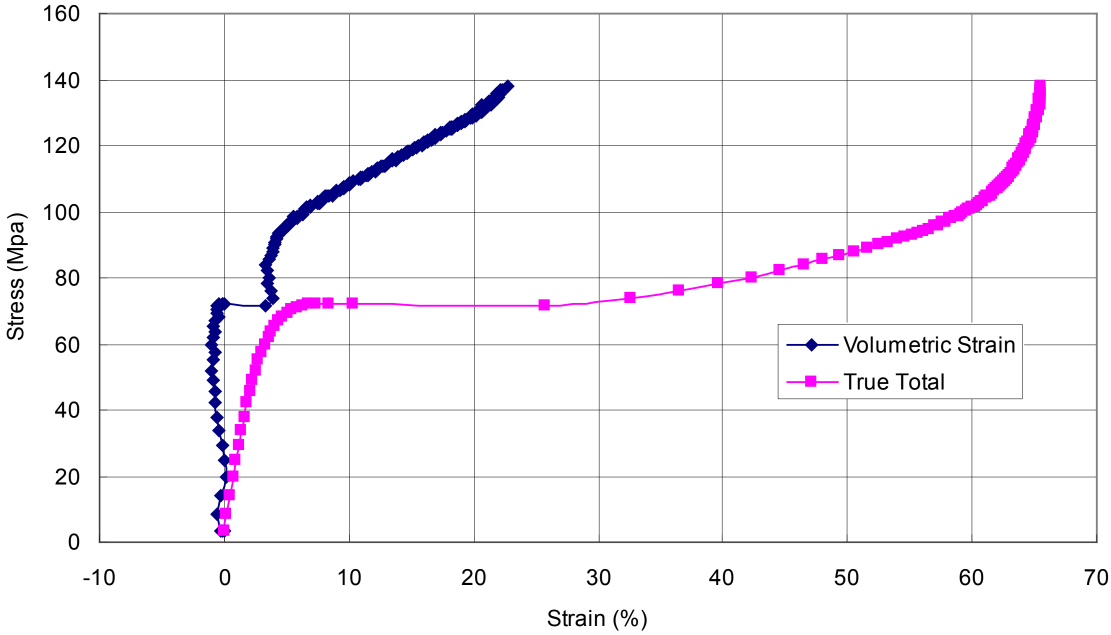

Fig. 4: Axial and volumetric strains v. true stress

We computed the total Poisson's ratio by using the ratio of local slopes of the axial (y) and transverse (x) strains. We observed that prior to yield, the total Poisson's ratio was 0.4,. At yield, there was a sudden drop in the total Poisson's ratio below 0.3 corresponding to the dramatic increase in volumetric strain. Knowing the Poisson's ratio in the elastic region at 0.4, the plastic Poisson's ratio could be calculated by subtraction from the total measured Poisson's ratio following the formulation proposed by Du Bois.

$$\upsilon_t \varepsilon_t = \upsilon_e \varepsilon_e + \upsilon_p \varepsilon_p \tag{1}$$

Where, subscripts t, e and p denote total, elastic and plastic respectively.

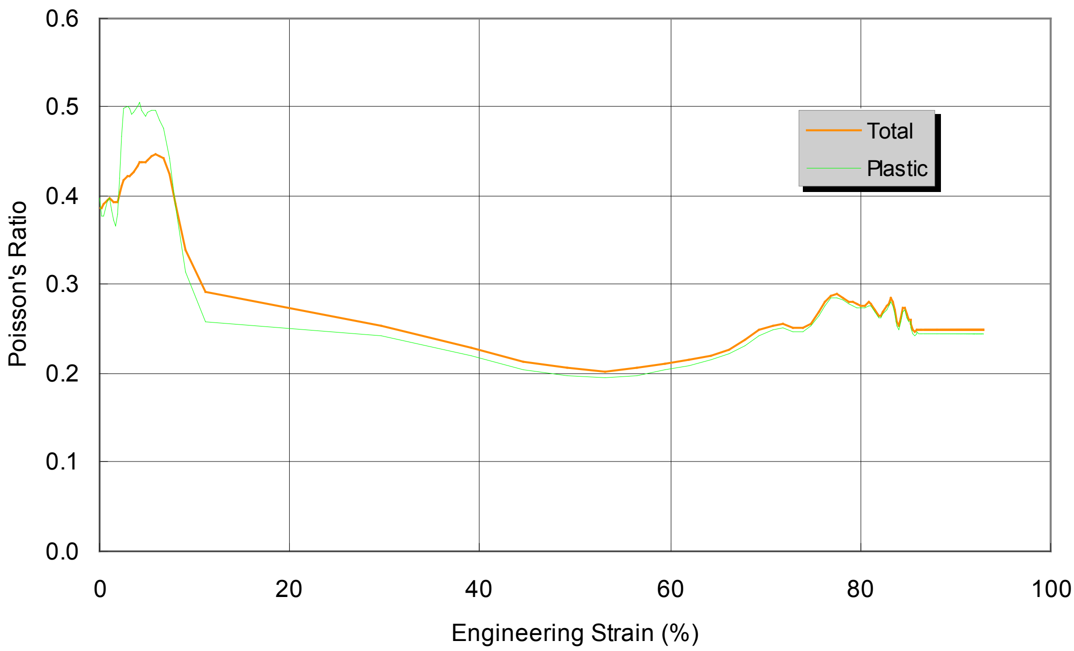

Fig. 5: Total and plastic Poisson's ratio with strain

As theoretically postulated, the plastic Poisson's ratio was found to be 0.5 prior to yield falling to about 0.25 post yield. It was interesting to note further, that the Poisson's ratio post yield remained relatively constant, possibly because the strain was localized with further straining occurring only in the localization region. This process then continued until failure.

In order to see how this data could be used with SAMP, we proceeded to compute true stress-strain from engineering stress-strain using the method proposed by Du Bois. It is clear from the data above, that the plastic Poisson's ration is less than 0.5. We first calculated the true stress strain curve based on classical equation below, setting the plastic Poisson's ratio to be 0.5.

$$\varepsilon_t = \ln(1+\varepsilon_e) \tag{2}$$

$$\sigma_t = \sigma_e (1+\varepsilon_e)^{2\upsilon_p} \tag{3}$$

Where suffixes t and e denotes true and engineering respectively while ν_p was the plastic Poisson's ratio. The resulting curve is shown in Figure 6 below. A region of negative slope was observed beyond the localization (necking point). Following Du Bois, a fictive curve was calculated by extrapolation using the equation below.

$$\sigma_t = \sigma * e^{2\upsilon_t(\varepsilon_t - \varepsilon *)} \tag{4}$$

Where $\sigma*$ and $\varepsilon*$ are the engineering stress and strain at localization, and ν_t , the total Poisson's ratio was supplied as a function of strain. There resulting curve is also shown in Figure 6.

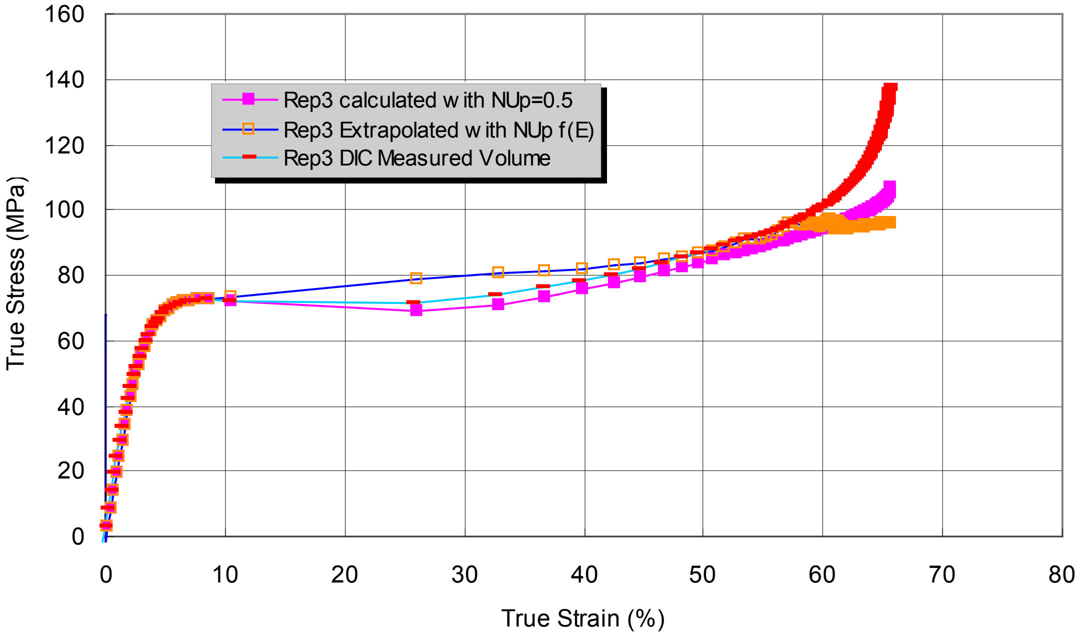

Fig. 6: Comparison of true stress-strain- direct measurement v. Du Bois calculation

We note particularly, that while the two curves bracket the DIC measured experimental data, they are deficient in not accounting for the fact that the Poisson's ratio related to the thickness of the material is not the same as that in the width direction in the post yield region. Correctly accounting for these values produces a stress-strain curve that is highly representative of the true behaviour of the material, requiring little or no corrections.

4 Determination of the Yield Locus

The yield locus was established by a number of quasi-static measurements in different stress states: tensile, biaxial, shear and compressive modes as described below. The tensile experiment was performed as described earlier. The compressive experiment was performed by taking a prism of the material 6.7 mm high (y), 12.5 mm wide (x) and 2.9 mm thick (z) and compressing it in the y-axis. A compressometer was used to ensure precise strain measurement. In an analogous experiment, for the biaxial tension data [3], a compressive test was performed in the thickness (z) direction of the material using a 25 mm diameter disk (y-x) , 2.9 mm thick (z). Lubrication was used to reduce the possibility of friction Strain was measured in the z-direction using a compressometer. The resulting z-strain was converted to biaxial tensile strain using the equation below.

$$\varepsilon_b = (1+\varepsilon_c)^{-0.5} - 1 \tag{5}$$

where suffix b and c refer to biaxial tensile and compressive respectively. The biaxial engineering stress was then computed using the following equation.

$$\sigma_b = \sigma_c / (1+\varepsilon_b)^3 \tag{6}$$

For the shear measurements, an Iosipescu fixture was used following ASTM D5379 [2], with the shear being measured in the x-y plane. Aramis DIC was used to perform the local strain measurements. We confirmed that the measurement was in a state of pure shear at yield. (Figure 7)

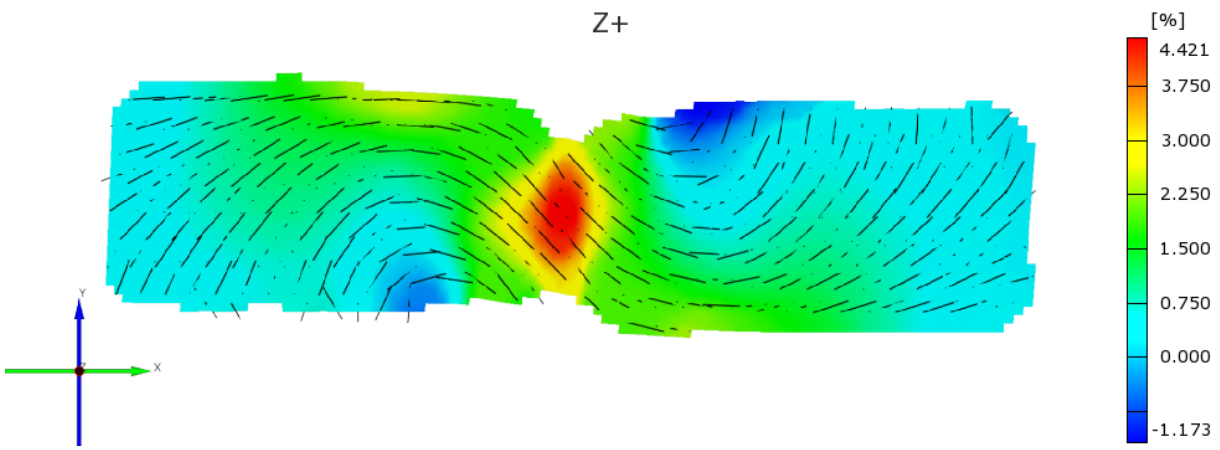

Fig. 7: Shear strain field at yield in the Iosipescu experiment

Taking the yield stresses in each mode and normalizing them against the tensile yield stress, we were able to plot the yield locus of the material. Figure (8).

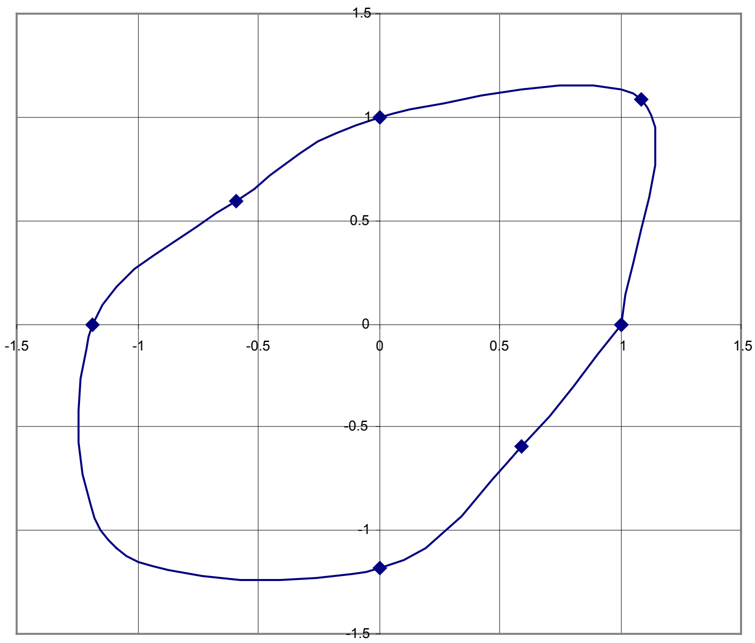

Fig. 8: Yield locus for polycarbonate

5 Summary

A suite of experiments have been described for generating material parameters required for SAMP. Good DIC measurements are shown to be of value in providing data for some of the more complex requirements that is free from experimental artefacts. Reasonably good measurement techniques have been proposed for determining the yield locus while at the same time, acknowledging that these measurements may not be entirely free from artefact. The DIC measurements themselves give valuable quantification of the volumetric and deviatoric components of polymer plasticity and strain evolution. Further, the material weakens post yield due to the increase in volumetric strain.

Note further that there is growing evidence that even in a tensile test, the polycarbonate material yields in a state of shear. In retrospect, this is not extremely surprising but merely that materials are weakest in the shear mode and also suggests that there may be harmonization of the post yield phenomenon as being initiated in shear and propagating in tension.

6 Literature

[1] Du Bois P.A., et al. “A Constitutive Formulation for Polymers Subjected to High Strain Rates” LS-Dyna International Conference Proceedings (2006)

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[3] Gese H., private communication (2013)

[4] McMaster Carr #85805K26 polycarbonate http://www.mcmaster.com/#catalog/119/3553/=mh95r8 (2013)

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[6] Diehl T.,“On the Measurement and Modeling of Polymeric Materials that Exhibit Unstable Localized Necking“, Diehl, T., Abaqus Users' Conference Proceedings (2007).