

A ROBUST METHODOLOGY TO CALIBRATE CRASH MATERIAL MODELS FOR POLYMERS

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Hubert Lobo and Brian Croop

DatapointLabs, Ithaca, NY, USA

THEME

Modelling of Materials

KEYWORDS

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SUMMARY

High strain rate material modelling for use in crash and drop testing has been plagued by a number of problems. These include poor quality data, material models unsuited to polymer behaviour and unclear material model calibration guidelines, to name just a few. This has rendered the modelling of polymers to be a risky proposition with a highly variable success rate. In previous work, we tackled each of the above problems individually.

First, we developed a consistent approach to obtaining clean, high quality tensile data on polymers at high strain rates. We noted that polymers exhibit behaviours that are unlike those of metals; further that we could divide them into groups based on particular behavioural characteristics. For example, while some polymers exhibited a rate dependency of modulus; others showed ductile-brittle transitions at high strain rates. A variety of post-yield behaviours were noted depending on polymer type and the presence of fillers.

Having a clear picture of the observed behaviour, we were then able to identify the limitations that existed in the crash material models that are in use today. We then developed guidelines for the selection of the right material model that best described the various kinds of behaviours exhibited by different classes of polymers. The calibration itself was found to depend on a series of pragmatic choices in order to best fit the complex observed behaviour to the simplistic material models available.

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To confirm our choice of material model and the calibration itself, we proceeded to validate the material model using computer simulation. High strain rate simulations of the actual tensile experiment were carried out at different strain rates. By comparing the simulation to experiment, we obtained a quantitative measure of the fidelity of the material model to the actual properties. This last step is essential to assure the users of such data that the material behaviour is being properly described in the more complex end-use simulations that follow.

1: Introduction

High strain rate material modelling of polymers for use in crash and drop testing has been plagued by a number of problems. These include poor quality and noisy data, material models unsuited to polymer behaviour and unclear material model calibration guidelines. The modelling of polymers is thus a risky proposition with a highly variable success rate. In previous work, we tackled each of the above problems individually. In this paper, we summarize and then proceed to present a material modelling strategy that can be applied for a wide variety of polymers.

First, we present a consistent approach to obtaining clean, high quality tensile data on polymers at high strain rates. We then identify the limitations that exist in crash material models that are in use today. We then present guidelines for the selection of the right material model that best describes the various kinds of behaviours exhibited by different classes of polymers. The calibration itself is found to depend on a series of pragmatic choices in order to best fit the complex observed behaviour to the simplistic material models available.

To confirm our choice of material model and the calibration itself, we then proceed to validate the material model using computer simulation. By comparing the simulation to experiment, we obtain a quantitative measure of the fidelity of the material model to the actual properties. This last step is essential to assure the analyst that the material behaviour is being properly described in the more complex end-use simulations to follow.

While the material modelling concepts presented here use the LS- DYNA software terminology, it is possible to translate these comments to other software codes that use analogous material models.

2: Measuring high strain rate properties

Measuring high-strain rate properties of polymers is a difficult proposition for a number of reasons. At high strain rates, noise and vibration appear in the data in the critical initial strain region, obscuring efforts to measure an accurate modulus, and potentially impacting even the yield stress value. Additionally, it is important

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to ensure that the desired strain rate is achieved and that the load frame is not in transition during the measurement. Slack adapters have been used in the past for this purpose. Data acquisition needs to be adequate so that enough datapoints are available for a quality measurement. Extensometry is very difficult at very high strain rates. Noise can present such a problem as to make the data useless for practical purposes.

Our approach seeks to combine a knowledge of polymers with the known limitations of test instruments to create a series of tests that we know will yield clean and dependable data. These factors are discussed below.

3: Choosing test specimens

The specimen choice is an important factor in obtaining quality data. The ASTM Type V specimen has been found to be highly effective in that it is small enough, that high strain rates can be attained with ease, yet it is representative of the behaviour of the polymer in general. In previous work [4], we showed that these test specimens yielded identical data to the conventional ASTM Type I specimen. These specimens can be derived by moulding the actual dog bones or by CNC machining from plaques. The small size also makes it feasible to obtain specimens from actual in-the-field moulded parts. Important factors that govern the source of test specimens include the type of material and the method of processing. Unfilled polymers that have no significant melt elasticity can be moulded with a fair amount of confidence. Polymers with high melt elasticity such as blow moulding grades used in bottles can experience significant alteration in properties when they are injection moulded, due to the strong orientation flows that occur in tensile bar geometries. Consequently, it is preferable to machine test specimens from flat panels in the blow moulded or thermoformed product. Thickness variation could be a problem. If moulded specimens must be used, a comparison between formed and moulded samples at quasi-static strain rates should be performed to quantify any differences.

Fiber filled materials present an additional degree of complexity with respect to stress-strain behaviour. While the analysis of the data will be treated later, it is important to note that orientation effects can be significant. Taking data from moulded tensile bars will invariably give high values. A compromise technique is to machine specimens in flow and cross-flow directions from long injection-moulded plaques. Simulation challenges in the use of this data will remain. New software developments have been presented to solve the problem, where bi-directional rate dependent data is scaled using localised fibre orientations obtained from injection-moulding simulation.

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Finally, there are concerns about taking test specimens from a moulded part. As stated earlier, the ASTM Type V geometry is well suited to the purpose because it can be taken from small flat regions in the part. Moulded parts however tend to have complicated geometry and this can cause a spatial variation of properties. Even though this variation may be small, the taking of test specimens from different locations in the sample part will cause the introduction of variability into the test data. This muddies the data making it much harder to create a robust material model. Should the luxury exist, test specimens should be taken from the same location; this requires a large number of sample parts.

4: The importance of capturing rate dependency in test data

Most impact situations will result in the component under study being exposed to a wide range of strain rates. For the material model to operate effectively, it is important for the model to be correct over this range. The effect of strain rate is experienced on a log scale rather than a linear scale. In other words, a doubling or tripling of strain rates is not an appropriate expression of the rate dependency. An adequate range would span 3-5 decades of strain rate. This kind of data also allows us to exploit some of the laws of rate dependent behaviour to ensure the validity of the material model, as we will explain later. The use of a single high strain-rate curve for impact simulation is inadvisable because accuracy decreases at high strain rate plus there is no means to apply a self-consistency check to verify accuracy.

It is useful at this point to also explain the difference between strain rate and impact velocity. Strain rate is a normalization of the impact velocity over a unit element, which, in the case of the test, is the gage length of the test specimen. An illustration of this concept follows: ASTM Type V specimens which are preferred for high strain work because of their small gauge length of 10mm allow five times higher strain rates to be reached using the same speed than a ASTM Type I specimen which has a gauge length of 50 mm. The large tensile bar experiences a lower strain rate. To gauge the relevant range of strain rates, it is useful perform a sample simulation with generic data and examine the results. The measured rate dependent data should try to cover the range of strain rates observed in the simulation.

5: Rate dependency of polymers

Polymers are highly complex materials, whose mechanical properties vary with stress level, time (rate), temperature and many other parameters. The result is non-linear behaviour that is not easily captured by conventional material models which

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have their roots in metals theory. The following kinds of effects must be considered.

Dependency of the stress-strain relation on stress level is unique for polymers. Hyper-elastic materials exhibit highly non-linear elastic behaviour but show no plasticity. Metals on the other hand, show highly linear elastic behaviour, with plasticity becoming relevant only post-yield. Polymer stress-strain behaviour is neither hyper-elastic nor linear. Contrary to metals, the onset of plastic strain occurs prior to yield. Additionally, the elastic behaviour is non-linear. Attempts to approximate this behaviour using metals theory comprise a poor approximation of the actual behaviour leading to the following compromises. Attempting to predict behaviour with fidelity to the onset of true plastic behaviour causes an under-prediction of the stiffness of the material at low stresses. Attempting to be true to the elastic modulus of the material results in an over-prediction of the plastic strain as one is forced to assume the onset of plastic strain much before it occurs in real life. The consequences and modelling of this behaviour are discussed in more detail in the paper by Lobo and Hurtado [5].

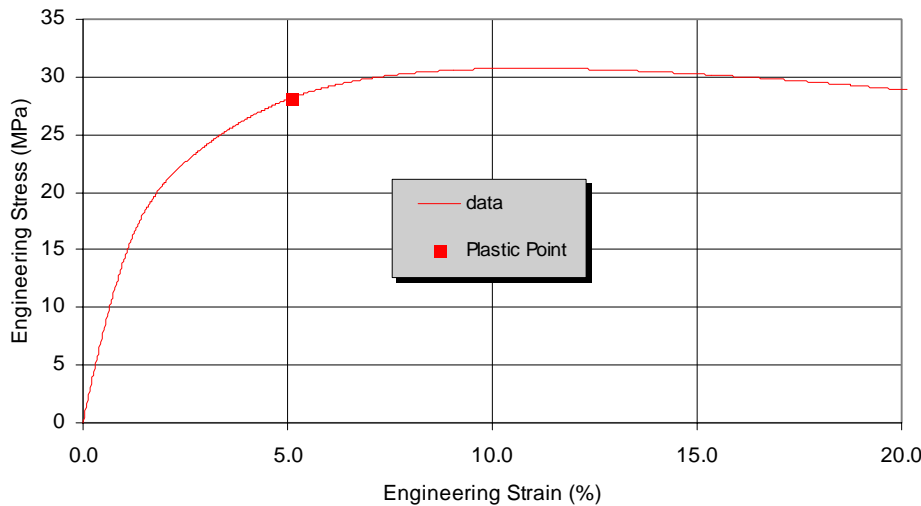


Figure 1 Non-linear behaviour of polymers

In describing the rate-dependent behaviour of a polymer, additional complications arise. Up to the vicinity of yield, we note that certain polymers exhibit significant rate-dependency of modulus while others do not [3]. This is in distinct contrast to metal behaviour where the expected behavioural trend is

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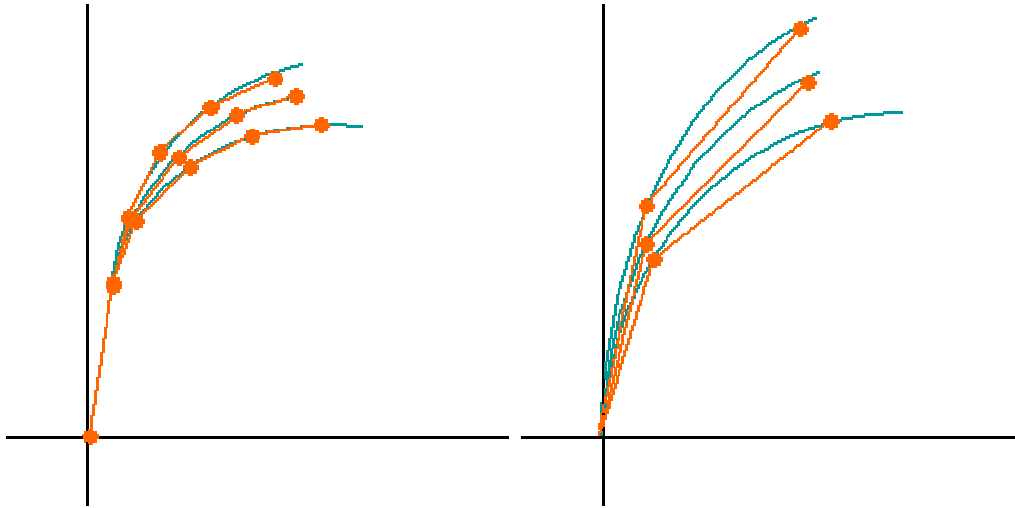


Figure 2 Modulus of polymers can be rate-independent or rate-dependent

toward no dependency of modulus with strain rate, as exemplified by the LS-DYNA MAT24 type material model. A consequence of this finding is that polymers exhibiting rate dependency of modulus cannot be described by a MAT24 type model. The use of a MAT24 type model for such materials will result in significant error in stiffness predictions. These and other limitations must be considered carefully in selection of a material model for describing rate dependency of polymers.

The combination of the effect of stress as well as strain-rate on the rate dependent stress-strain relationship of polymers, as explained above creates a complex situation that is only crudely approximated with currently available material models. Nonetheless, by proper selection, is possible to conduct meaningful simulation by selecting existing material models that most closely match the behaviour shown by the material data.

With respect to the rate-dependency of the plasticity behaviour, a remarkable consistency is observed for a large variety of polymers. A predominant trend exists toward agreement with the Eyring equation, which is characterized as a linearly

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increasing relationship between yield stress σ and log strain rate. The

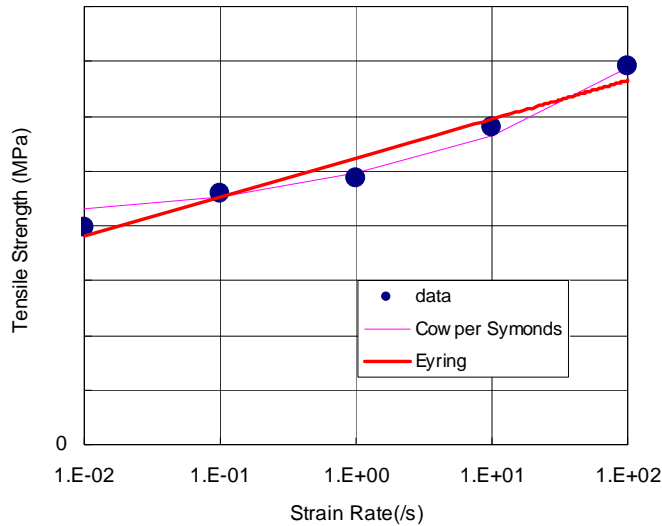


Figure 3 The Eyring v. Cowper Symonds equations for rate dependency of yield stress

obvious exception is the case of polymers exhibiting brittle failure where the result is more noisy. In contrast, the Cowper-Symonds model which is used extensively for metals and is implemented in MAT24 fails to capture the correct trend leading to modelling inaccuracy in modelling rate dependency of polymers.

The addition of fiber reinforcements to polymers is a common practice to increase the strength of these materials. In addition to an increase in stiffness, the nature of the failure changes when fillers are added. In extreme cases, such as highly glass filled polymers, the failure changes from ductile to brittle. Interestingly, with intermediate filler loadings, there is a gradual change from ductile to brittle failure with the increase in strain rate. This variation in post-yield behavior with strain rate is not easily captured in available material models today.

6: Material model calibration for ductile polymers

The LS-DYNA MAT24 is the most widespread material model in use today for the modelling of crash, drop and other rate-dependent phenomena. It's simplest and most commonly used nuance couples a Cowper-Symonds model with an elastic-plastic curve as follows. The elastic region is modelled as rate independent up to an arbitrarily or otherwise determined yield point, beyond which the stress-strain curve at the lowest strain rate of interest is decomposed into an elastic-plastic model. This produces a curve of stress v. plastic strain (ES v. EPS),

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hereafter referred to as the plasticity curve. As seen in Figure 2a, the definition of yield as seen in metals does not correspond to that used in this context for polymers. The accuracy of this model, when applied to polymers, depends on: the stress-strain relationship being linear up to the chosen yield point; that this initial linearity is not rate dependent; and that the shape of the plasticity curve is uniform and independent of strain rate. This is simply not true for most polymers. Since polymers are non-linear elastic followed by elastic-plastic, an arbitrary choice is usually made somewhere along the increasing part of the stress-strain curve denoting the onset of plastic strain limitation of linear elasticity.

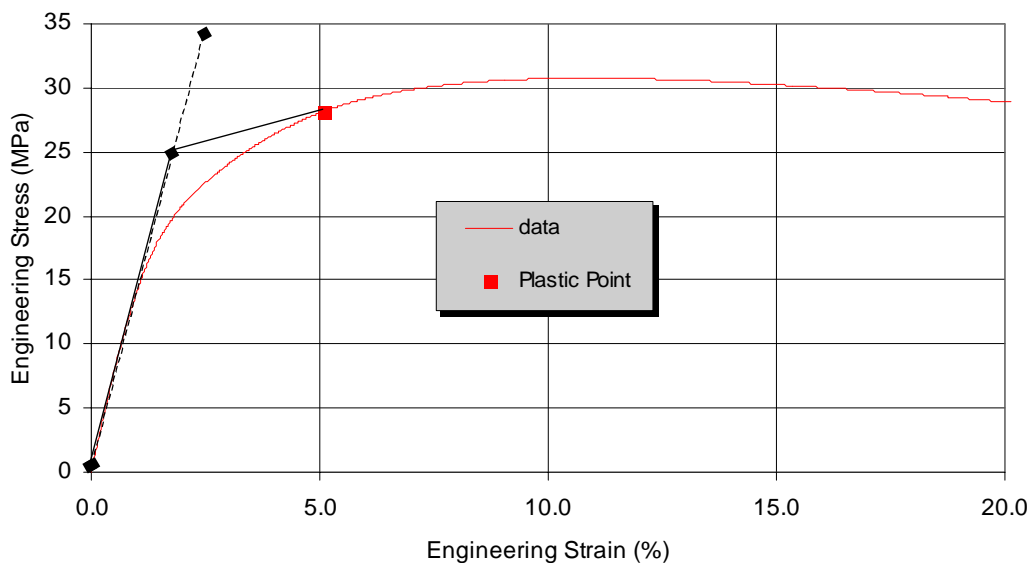


Figure 4 Selection of the initial region of the MAT24 model

In recent work, we showed a method for accurately measuring this "plastic point" [5]. Using the plastic point, however is not always feasible because of non-linear elastic behaviour. As illustrated in Figure 4, fidelity to the linear elastic region results in an over-prediction of plastic strain because the material continues to be elastic at stresses far exceeding the "linear-elastic" region. On the other hand, using a secant modulus to describe the behaviour up to the plastic point results in a material model that significantly under-predicts the stiffness of the material in the elastic region. Currently, with the the MAT24 model, there is no recourse other than to choose, pragmatically, a plastic point that is somewhere in between these extremes, using the initial elastic modulus for EMOD instead of the secant modulus and then picking a yield point at a stress somewhat below the plastic point stress for the von Mises yield.

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Once EMOD has been chosen, it is a simple matter to discretize the static stress-strain and convert the data into plastic strains following the normal rules of the elastic plastic model. Applying the Cowper-Symonds equation, it is now possible to scale this curve to other strain rates. The equation has the advantage of smooth extrapolation without limits. However, since the equation is incapable of truly describing the rate dependency of the yield phenomenon (Figure 3), it cannot accurately scale the plasticity curve to high strain rates. A possible solution is to use the LCSR option, which permits the submission of a table of scale factors for each strain rate. LCSR is an interesting option which allows fidelity to the test data. However, it must be used with caution. High strain rate data is experimentally difficult to obtain so that there is often scatter in the data. This scatter must be smoothed in some way so that the resultant model contains no spurious behaviour. Since we know that the Eyring Equation appears to accurately describe the rate dependency of most ductile polymers, the LCSR table can be derived from a best fit of the yield stress v. log strain rate data. This approach carries two advantages: first, the elimination of noise and second, the ability to extrapolate the model to 'higher than tested' strain rates, since LCSR based MAT24 terminates rate dependency computation when the highest strain rate in the table is exceeded. Using MAT24 with LCSR as described above, we can successfully overcome the limitation of the Cowper-Symonds model in the simulation of polymer rate-dependency.

A serious limitation of MAT24 arises from the fact that the actual failure strains typically decrease with increasing strain rate. This variation is not accommodated by the model, which assumes that failure strain is constant and independent of strain rate, as would be typical for metals. Failure in MAT_24 occurs when the accumulated plastic strain in an element reaches the failure stress value specified in the FAIL term. At each time step, after the trial stress is computed, if the trial stress is found to be outside the yield surface (VonMises), LS-DYNA scales the stress back to the yield surface and then obtains the accumulated plastic strain by using the material model to calculate the corresponding effective plastic strain (EPS) at the strain rate seen by the element. If this accumulated plastic strain equals FAIL, the element is removed. FAIL is usually chosen by the analyst as the largest failure strain in the material data. This is the conservative approach. If the data shows a variation in failure strains with strain rates, a check must be made by the analyst to review the strain-rate experienced by the part, to pick a value of FAIL at that corresponding strain rate. This is described later. It is clear that with polymers with ductile-brittle transitions or where the failure strain is highly rate dependent that this limitation can have a significant impact on the simulation.

The LCSS option of MAT24 is very useful when the shape of the plasticity curve changes with strain rate. This phenomenon is often observed in polymers. In this

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case, by submitting a plasticity curve for each strain-rate, we are able to independently describe the stress-strain behaviour as a function of strain rate allowing us the ultimate in flexibility offered by the model. It may still be a useful exercise to smooth the rate-dependency using the approach outlined earlier. LCSS however does not offer relief in the modelling of ductile-brittle transitions, because of the limitation of FAIL. Proper implementation of LCSS requires that we extrapolate all the plasticity curves to the largest failure strain (FAIL) that we intend to use in the model. Consequently information regarding the change in failure strain with strain rate is lost.

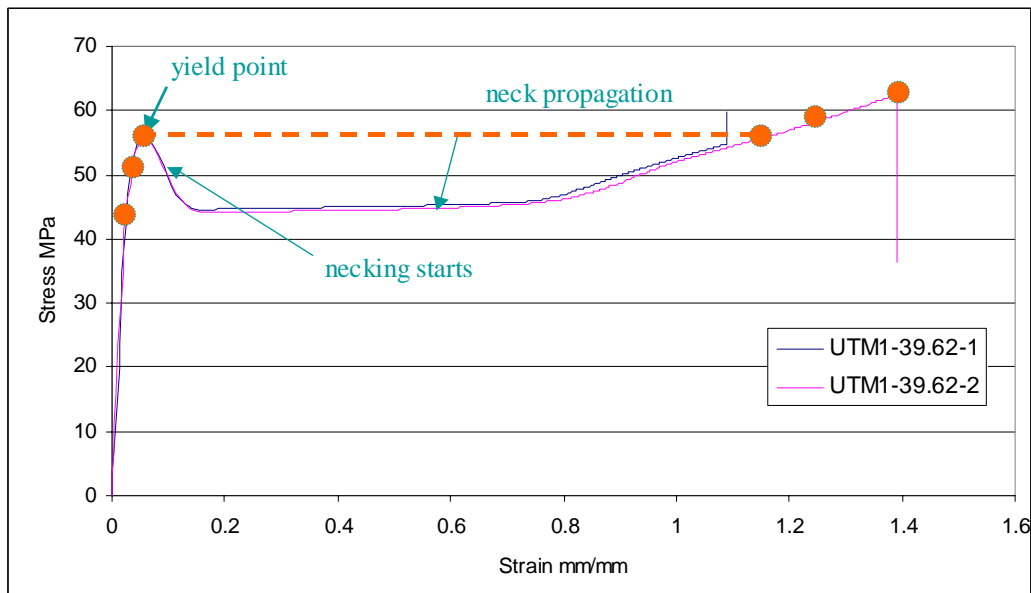


Figure 5 Point selection for polymers with necking behavior

Polymers such as polycarbonate, polyethylene and polypropylene exhibit long tails of post-yield strain and are capable of absorbing significant energy in this phase of their deformation. Stress-strain curves for polymers that are not brittle in nature go through an inflection or local maximum commonly referred to as the yield point, not to be confused with the von Mises yield which corresponds to the onset of plastic deformation. When handling post-yield behaviour, a number of complications arise. Most post yield behaviour is accompanied by necking, localized non-uniform deformation, which leads to a condition where the cross-sectional area of the deformation zone is unknown. Consequently, the stress is also an unknown and can only be crudely estimated by making some assumptions about the cross-sectional area. The most common procedure is to assume that the true stress calculation applies in this region as well. A consequence of this assumption is that the slope of the stress-strain curve is seen to gradually increase

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with increasing strain. In the case of olefinic materials such as polypropylene and polyethylene whose necking phenomenon is more closely equated with unravelling or unzipping of the dendritic structure, it is more likely that the stress remains constant during necking. In any case, to model these regions using MAT24, it is only essential to eliminate negative slopes in the model as illustrated in Figure 5.

7: Models for Fiber-filled and Brittle polymers

A number of fiber-filled polymers show considerable rate dependency of modulus followed by very small strains to failure. Yield and failure points are scattered. A small amount of plastic strain is accumulated prior to failure. This kind of behaviour is extremely difficult to model with accuracy using MAT24 for the following reasons. The stress-strain curves begin to diverge almost immediately as seen in Figure 2b. Consequently, MAT24 will either under-predict the stiffness at low strain rates or vice versa depending on the choice of the modulus E . While this may have quite a bit of impact on the simulation of most polymers, in the case of filled polymers, the effect is even more dramatic because the failure strains are so small, typically 2%. MAT 19, even though it suffers from the deficiency of being a bi-linear model, is better suited and achieves higher fidelity to the experimental data. An added advantage of the model is its ability to precisely indicate the failure envelope of the material via the use of the failure strain v. strain rate dependency option. This effect is often quite marked in such materials and must be modelled with accuracy for good simulation results. Additionally, the model allows failure based on tensile plastic strain only.

8: Modelling Ductile-Brittle transitions in polymers

MAT89 is an elastic-plastic material model that does not require the user to break up data into elastic and elastic-plastic regions. It is recommended by LS-DYNA to handle the highly complex behaviour of ductile-brittle transitions where failure strains can vary anywhere between 100% and 10% for some polymers. With MAT89, the initial stress-strain curve is entered as true stress-strain data. LS-DYNA internally checks the slope of the curve. When this slope falls below the modulus E specified in the material card, the material is assumed to have yielded. The treatment of plasticity then follows MAT24, as described earlier. The LCSR scaling of the stress-strain curve allows this model to be scaled to high strain rates in a manner similar to MAT24. The LCSR table of yield stress v. strain rate is a better choice for modelling rate-dependency than the Cowper-Symonds equation for the same reasons described earlier. The key benefit of MAT89 is the LCFAIL table which enables the entry of failure strains v. strain rate. This feature

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overcomes the limitation of MAT24, which restricts its ability to model polymers whose failure strains change significantly with strain rate.

True Tensile Stress-Strain Curves

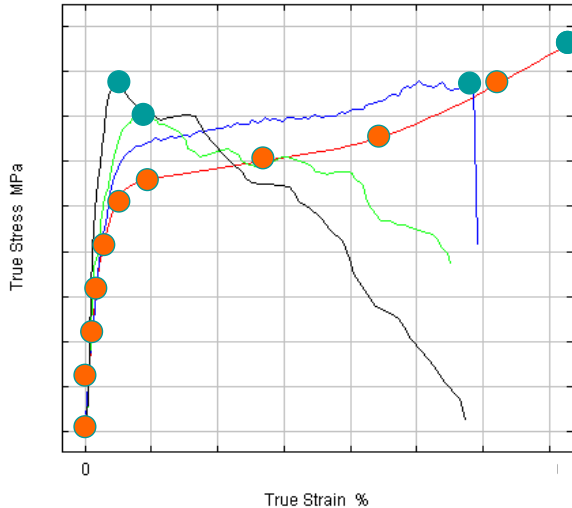


Figure 6 Blue dots represent the use off LCFAIL in MAT89 for failure modelling

9: Validation of Material Models

A number of variables exist in the material model calibration process that can have a considerable impact on how the simulation sees the material data. The reasons for this have been outlined in this paper, ranging from poor material data to an improper choice of material model to an incorrect assignment of any number of model variables. To introduce some measure of confidence in the calibration process, a simple simulation of the actual experiment is quite invaluable. The ASTM Type V was modelled in LS-DYNA. The model had nodes at the extensometer locations in order to extract the strain. The model was subject to two impact velocities, one slow and one fast. The stress-strain curve obtained from the simulation was compared to that obtained from the original experiment. Figure 7 illustrates such a comparison for a polymer at a strain rate of 10/s. The material model data at 10/s converted back to the stress-strain form is presented for reference purpose. The failure strain (FAIL) was set to 0.25 by trial and error, to force the failure point to coincide with the experimentally observed failure at 10/s. This trial and error process actually provides a means to set a meaningful failure strain that bears some semblance to reality. While the validation may not be a predictor of simulation accuracy for a complex part, it does ensure that an improper material calibration is not a reason for a poor real life correlation. While the model may be quite effective in predicting deformation, the more complex

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forms of failure from multi-axial stress states may not be well handled without additional testing. Such steps could include tuning the model using a multi-axial impact experiment such as falling dart impact or a triaxial impact, as would be obtained from a notched Charpy or Izod experiment.

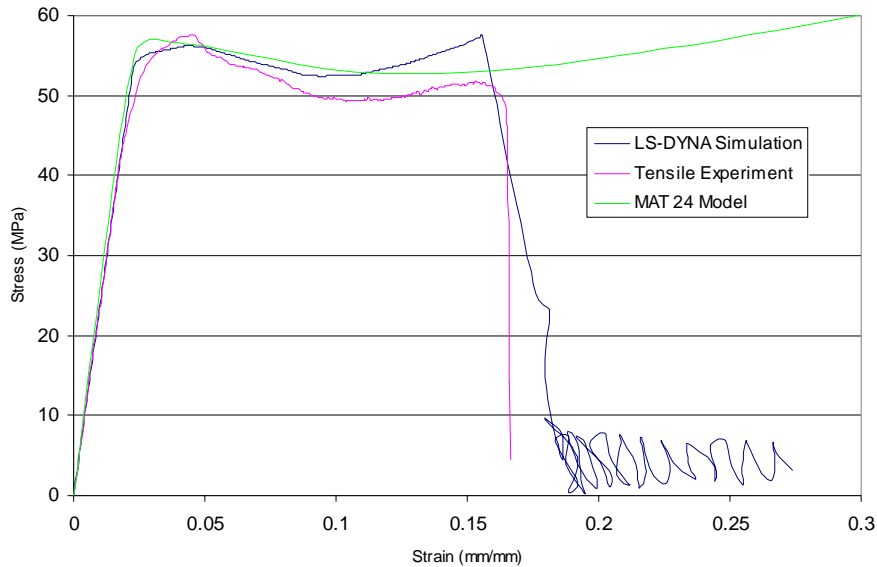


Figure 7 Validation of the simple tensile experiment

10: Conclusion

A number of material models exist that could be used for the modelling of polymeric behaviour. Sophisticated new models have been proposed recently and their calibration procedures are described in detail elsewhere. However, their use is not commonplace in the world today for many reasons such as unfamiliarity, complex model calibration and a slow down in simulation performance.

This paper provides a methodology for the selection and use of commonplace material models based on the observed high strain-rate behaviour of a polymer. Constraints in these models limit their ability to capture all the possible behaviour types of polymers. It is important, therefore, to select the appropriate material model in order to exploit its capabilities to the maximum. There still remain important behaviours that are not adequately captured by any model. A simple validation permits the analyst to assess the fidelity of the material calibration. Such a validated material model can form a basis for material model tuning based on more complex experiments.

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11: Acknowledgements

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