INFLUENCE OF SPECIMEN ORIENTATION ON THE PLASTIC INSTABILITY AND NECKING OF EXTRUDED HDPE SHEET

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الملخص

يتضمن هذا العمل دراسة عملية وتحليلية لبدء وتطور التشكيل اللدن اللامستقر (Plastic Instability) لبوليمر البولي ايثيلين العالي الكثافة (HDPE). سلسلة من تجارب الشد أحادية المحور باستخدام عدة مسارات تحميل عند مستويات انفعال كبيرة تم إجراؤها وذلك من أجل دراسة تأثير كل من معدلات الانفعال المختلفة واتجاه عينات الشد المصنعة من صفيحة من البولي ايثيلين المبثوق على السلوك الميكانيكي (Elasto-viscoplastic behavior) خلال مرحلة التعنق (Necking). كما وتضمن هذا العمل متابعة ودراسة للتغير الحادث في خصائص المرونة والمتمثلة في (معاملي يونج وبواسون) وتطورهما نتيجة لظاهرة الانهيار الحادث أثناء مرحلة التشكيل اللدن اللامستقر. إعتمادا على معيار كونسيدير (Considère Criterion)، تم إقتراح إجراء بعض التعديلات عليها والتي تهدف إلى الأخذ في الاعتبار التغير في الحجم الحادث عند تشكيل المواد البوليميرية. المقارنة بين النتائج النظرية المتحصل عليها من النموذج المعدل وتلك المتحصل عليها من التجارب العملية تبين مدى قدرة النموذج على التنبؤ ببداية واستقرار ظاهرة التعنق (1st and 2nd Considère Strain)، كما ونوقش هذا النموذج المعدل ايضا من ناحية تأثير الاتجاه ومعدلات الانفعال المختلفة.

ABSTRACT

In this work, the initiation and stabilization of the Plastic Instability of High-Density Polyethylene have been experimentally investigated and analyzed. Tensile tests using several complex loading paths involving large deformation have been carried out in order to determine the effect of different constant strain rates and various specimen orientations of extruded HDPE sheet on the elasto-viscoplastic behavior during necking. This work also examines the evolution of Poisson's ratio and Modulus of elasticity of semicrystalline polymers due to phenomenon of damage during instability stage. Based on the Considère criterion, a modified instability criterion is proposed taking into account the significant effect of volume variation due to damage under large deformation. The ability of the modified criterion to predict 1st and 2nd Considère strain of this polymer is examined. A good agreement between the modified criterion and the available experimental data is attained. This criterion is also discussed in relation to the orientation behavior and strain rate.

KEYWORDS: High Density Polyethylene; Tensile Necking; Plastic Instability Criterion; Orientation Dependency; Mechanical Behavior.

INTRODUCTION

Under uniaxial tension, macroscopic deformation of polymers is generally associated with localization phenomena due to inhomogeneous plastic deformation. Unlike metals, this plastic instability does not lead to failure, but it leads to the process of neck formation, stabilization and propagation. The mechanical properties are improved during plastic deformation due to reorientation and the new microstructure that is created in the material during the deformation process [1-4]. Therefore, the stretching behavior of semicrystalline polymer was the subject of many scientific researches attempting to explain the necking phenomena in physical terms and aiming to predict and optimize the neck formation and propagation and to study the parameters which control these processes [5-10].

Elasto-viscoplastic behavior of semicrystalline polymers is strongly dependent on their complex microstructure with the presence of both crystalline and amorphous phases which make them interesting materials for many studies. Morphological parameters, strain rate, temperature, thermal aging, and stress triaxiality ratio are among the most important factors which have a strong influence on nonlinear behavior of these materials, and which are the subject of several experimental and theoretical studies developed on a wide range of polymers [3,11-16]. Despite the large influence of damage during plastic instability on the elastic modulus and Poisson's ratio, some attention has been devoted to characterize the evolution of elastic properties during the development and neck propagation [17-19].

Anisotropic behavior displayed during deformation in semicrystalline polymers have a scientific effect on their mechanical behavior. Although this particular property was investigated through tensile tests by study the specimen orientations effect on mechanical properties and morphology [20-23], a very limited studies revealed the influence of specimen orientation and strain rate on neck formation and propagation during plastic instability of extruded polymeric sheets.

The aim of this study is to experimentally investigate and analyze the effect of the different strain rates and various specimen orientations of extruded High-Density Polyethylene on the elasto-viscoplastic behavior during the initiation and propagation of plastic instability. The evaluation and dependency of Poisson's ratio and Modulus of elasticity due to phenomenon of damage during necking has been also measured. Neck formation is generally interpreted with Considère criterion which has been applied to polymers at constant volume. Therefore, in next section, the objective is to revisit this most popular criterion of plastic instability in order to provide a modified instability criterion which is able to take into account the volume variation due to damage under large deformation of semicrystalline polymers.

MATERIAL AND EXPERIMENTAL METHODS

Material

The High-Density Polyethylene, HDPE, employed in this study is a thermoplastic polymer. It is one of the most widely used in different industrial applications particularly in mechanical engineering, marine, and oil industry. It is reported to be very tough even at low temperatures and has good sliding properties and good processing and machining characteristics.

HDPE material is denoted as Polystone D by the supplier Röchling Group and it is provided in the form of 6 mm thick sheets produced by extrusion process. This material (under the same reference) has been studied in many scientific research [9,20,24]. The experimentally measured density, crystallinity ratio, glass transition temperature and melting point, and other additional data for the material are listed in Table (1).

Table 1: Characteristics of HDPE used in this study.

Density	0.952 g/cm^3
Crystallinity ratio	66 %
Glass transition temperature	-125 ˚C
Melting point	135˚C
Molecular weight	500,000 g/mol
Service temperature range	-100 to 80˚C
Modulus of elasticity	1200 N/mm^2
Yield stress	27 N/mm^2

Specimens Geometry

The tensile test specimens were prepared according to the geometrical configuration shown in Figure (1). This specific geometry has been designed in order to control the appearance of necking and localize the plastic deformation in the central part of the specimen; therefore any change of the cross-sectional dimensions can be easily monitored and determined. The specimens were cutted and machined from extruded sheet in three angles 0˚, 45˚ and 90˚, measured from the extrusion direction. One must take into account in which orientation the transverse strains ε_2 and ε_3 are equal or not equal, which represent the isotropic or anisotropic cases. Before any mechanical tests, specimens were submitted to annealing heat treatment for 24 hours at 120 °C in vacuum, then cooled slowly to obtain a microstructural state without residual stresses [25].

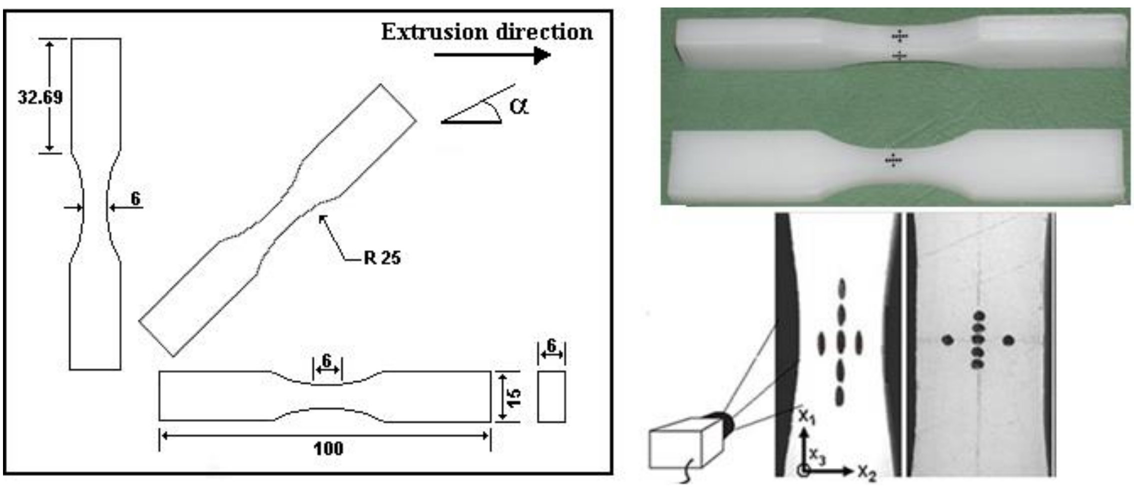

Figure 1: Geometry of tensile specimen (dimensions in mm) and determination of principal true strain in the RVE using 2*7 dot markers.

Experimental Setup

The tensile tests were performed at controlled room temperature using MTS 810 universal testing machine, equipped with an optical measurement system called videotraction extensometer that controls the strain rate and measures the local true strains. The details of this technique were previously published in [8,26]. In order to avoid the highly questionable assumption of transversally isotropic, which may lead to inaccurate estimation of volume change and Poisson's ratio, a double videotraction extensometer was chosen with 2x7 dot markers technique, in which the tensile specimens were printed with 7 dot markers on the main face, plan (1, 2), and 7 dot markers in the thickness, plan (1, 3), as in Figure (1). While force level F was measured by the load cell of the machine. CCD camera and image analysis software were used to determine the mechanical variables by analyzing the distances between dots aligned in both axial and transverse directions. This technique allows to determine the true strains in the three orthogonal directions, $\varepsilon_1, \varepsilon_2$, and ε_3, corresponding to the strain along the direction of the applied load, width, and thickness, respectively.

A particular attention was drawn on the necking phenomenon, which appeared in the central area of the specimen where the three dot markers were aligned perpendicular to the tensile axis (1). Therefore, one of the most important advantages of this non-contact method is that all measurements of true strains were done in real time in a representative volume element (RVE) situated at the center of the neck, which gives access to the volume change and Poisson's ratios whatever is the deformation paths in the necking region.

In order to quantify the influence of plastic instability during necking on the elasto-viscoplastic behavior particularly on the evolution of Poisson's ratio of HDPE, several loading paths mainly in uniaxial tension was applied up to large strain. These include a monotonic tensile test and Step-Cycle testing where the tensile specimen interrupted by unloading-reloading cycle at different strain levels. Each experiment was carried out with three to five series of tests under the same testing conditions. The reproducibility of the results was checked, and it was satisfactory.

RESULTS AND DISCUSSION

The experimental results shown in Figures (2 and 3) illustrate the effect of specimen's orientation, with different angles (0˚, 45˚ and 90˚) relative to the extrusion direction, on the mechanical behaviour of HDPE at large deformation under uniaxially stretching tests at room temperature and constant strain rate of $\dot{\varepsilon} = 0.001\ s^{-1}$. The photographs in Figure (3) show the evolution of deformation at the center of the specimen where the necking is initiated and propagated for the specimen oriented at angle 90˚.

The test data of the extruded HDPE show a planar anisotropic behavior for the three directions of specimens, this confirms with the results obtained by Rida et al and Laurent et al [27,28] using similar 6 mm thick extruded sheet of HDPE. It can be seen that, the transverse strain ε_2 and ε_3 are equal in the extrusion direction (angle 0˚), therefore, the material is transversally isotropic in this case, while it is anisotropic for the other directions (45˚ and 90˚), where the contraction of the width of specimen was larger than along of its thickness.

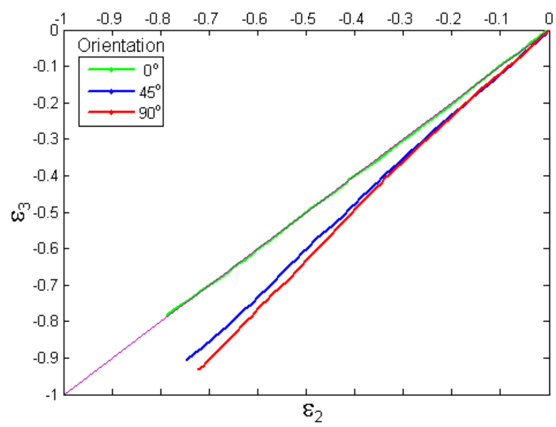

Figure 2: The anisotropy by means of transverse strains ε_2 and ε_3 as function of specimen orientation at strain rate of 0.001s^{-1}.

All three orientations display a similar typical force – true strain relationships with the same various domains (Figure 3) these evolutions show an orientation dependency (anisotropic effect) which are visible at large deformation. All force-strain curves reach a peak in terms of a maximum force just at yield point (close to $\varepsilon_1 = 0.11$). There is less than 3 % difference in the maximum force level obtained from the two directions 0˚ and 90˚. This peak is followed by a drop in the force level, the amount of the decrease in forces depends on the specimen orientation, this decrease reaches 27 % for extrusion direction, angle 0˚, while it is 38 % for angle 90˚. The specimen orientation affects also the level of axial strain at which the forces begin increase again, which reaches a 1.02, 1.15 and 1.27 for angles 0˚, 45˚ and 90˚ respectively.

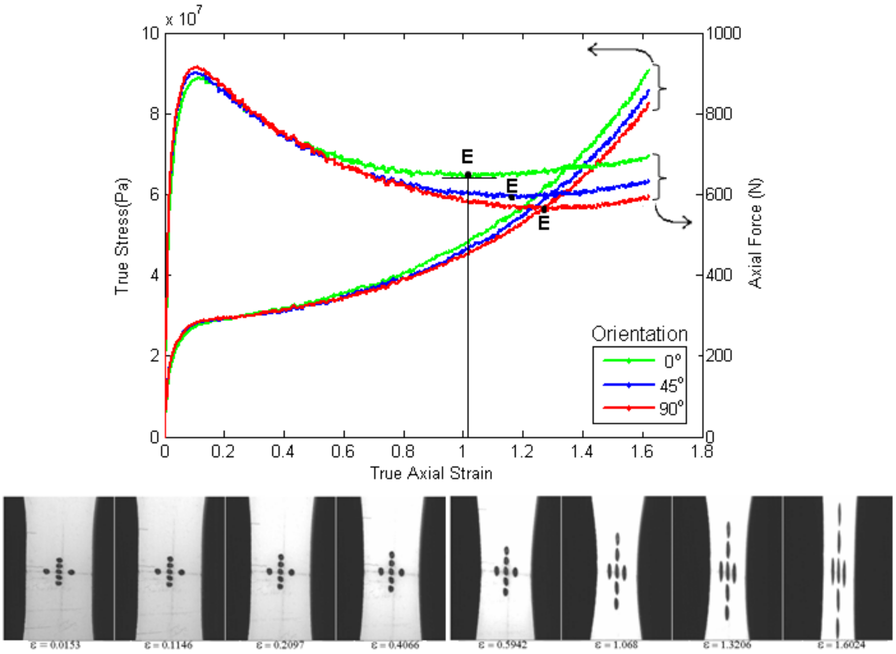

Figure 3: Evolution of axial force & true stress vs. true axial strain during monotonic tensile tests at strain rate of 0.001s^{-1}.

Figure (3) also shows the axial true stress plotted against axial true strain. A high nonlinear behavior of HDPE that characterized by four distinct regions is observed: elastic region, initial necking region, neck propagation region and strain hardening to fracture region. Based on the experimental results for different orientations, the general observed behavior are as follows:

1- Before the yield point at which the stress shows a maximum, the specimen is deformed homogeneously with limited linear elastic response at very low strains, and before reaching the yield point for a strain value close to 0.11 for all directions, the viscoelastic behavior takes place.

2- After the peak stress, the pseudo-plateau on the stress-strain curve is observed corresponding to the softening of the polymer, due to the more concentration of local strain in the specimen center. It is causing a decrease in cross-sectional area (rise of Poisson's effect), which leads to the instability effects induced by inhomogeneous deformation called neck initiation. This phenomenon can be related to the creation and development of microcavities in the amorphous phase and/or destruction of crystalline phase [1,8,16,29].

3- This neck becomes the nuclei for necking; the necked portion propagates towards both ends of the specimen, leading to transformation of the initial spherulitic; the crystalline lamellae will fragment into smaller blocks coupled by tie molecules and the fibrillar morphology will be developed [1,26,29].

4- After the neck propagation is completed over the entire specimen length, the stress rises again and therefore the strain hardening region is then apparent. The strain at this point (E in Figure 3), where the force is minimum, corresponds to the onset of necking stabilization known as natural draw ratio, which is normally used as a measure of material drawability. The natural draw ratio value depends on the specimen orientation, it increases as the angle increases. In this stage the spherulitic morphology is completely transformed into fibrillar structure with blocks and tie chains which becomes oriented in the direction of the tensile axis, causing the material to become stiffer in the axial direction [6,9,10,30,31].

Measurement of Elastic Properties Evolution

Among the mechanical properties of semicrystalline polymers, the damage has high influence on the elastic modulus and Poisson's ratio, as widely reported [17,27,32,33]. The evolution of elastic properties during plastic instability can be obtained by uniaxial step-cycle loading tests at several strain levels as shown in Figure (4) for the specimen orientation with angle 45˚. The elastic modulus of the damaged material is determined as the average slope of unloading-reloading cycles (line HD in Figure 4) which will be a valuable indicator of the damage, while the deformation can be admitted as reversible elastic along this line. This slop is not equal to the initial Young’s modulus and it defines the effective or apparent Young’s modulus (damaged modulus).

As necking progresses and propagates, the specimen will increase in length particularly in the center of necking due to longitudinal strain, and decrease in width and thickness due to lateral strains. As a result, there will be evolution of volumetric strains caused by necking. Poisson's ratio has a direct influence on the volume change of material, and its evolution can be obtained in the same way as above by measuring of ε_2 and ε_3 during the uniaxial loading-unloading-reloading test in the RVE where exactly the necking initiates and propagates.

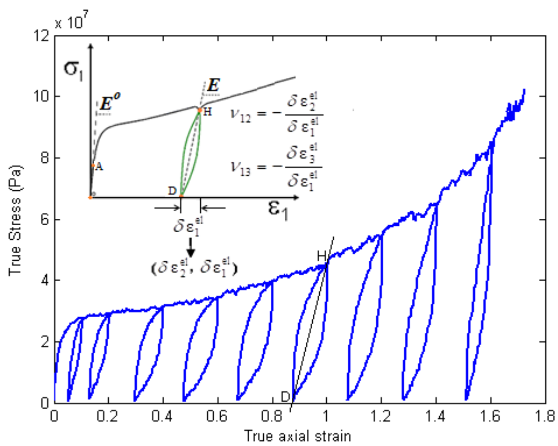

Figure 4: Measurement of apparent young’s modulus and Poisson's ratio represented by the line (DH) during uniaxial step-cycle tests for angle 45˚ and strain rate of 0.001s^{-1}.

Damage Effects

The Poisson's ratio in the necking region is strongly dependent on the specimen orientation as shown Figure (5). The Poisson's ratios, ν_{12} and ν_{13}, of all orientations have almost the same evolution but their amplitude is significantly different. Except for the isotropic case, angle 0˚, the other orientations show values of ν_{12} greater than ν_{13} due to the contraction of the width of specimen was larger than that of the thickness of specimen. It can also be noted that the increase in Poisson's ratios ν_{12} with the increase in orientation angle, was no longer applied for ν_{13} where the increase in orientation angle showed a decrease in Poisson's ratio, this may be due to the inhomogeneous deformation properties of the material. It is interesting to point out that, regardless of the specimen orientation, there is a volume change in this polymer [9,20], then the Poisson's ratios, ν_{12} and ν_{13}, may be higher than 0.5 indicating structural changes and instability in the material, as reported in the literature theoretically and experimentally [27,34-36]. Therefore, during necking, the evolution of elastic properties of HDPE is strongly related to the microstructural changes occurring during the plastic deformation. This complex phenomenon must be linked to all mechanisms of deformation and damage involved in the HDPE. The most important factors which can play a vital role in this evolution during the tensile deformation are: i) Creation and development of microcavities. ii) Spherulitic morphology change. iii) Reorientation of macromolecular chains, thus different stages can be enlightened from these curves:

1- The first stage corresponds to the elasto-viscoelastic behavior occurs at low deformation up to the yield stress. The Poisson's ratios showed a slight increase with the axial strain, which may be attributed to volume dilatation explained by elastic expansion and cavitation's [26, 29, 37]. A maximal value is reached at yield point, this maximum was highly influenced by the specimen orientation, and increases with the increase in orientation angle, for example ν_{12} reached approximately 0.47 at angle 0˚, 0.54 at angle 90˚.

2-The second stage corresponds to the necking initiation and propagation where the Poisson's ratios decrease rapidly in the beginning and then slowly until the strain levels close to 1.02, 1.15 and 1.27 depending on the specimen orientation, which are corresponding to the natural draw ratio. This may be attributed to the volume compaction of the amorphous phase due to rearrangement of lamellae orientation accompanied by the changing of voids shape and/or to the closure of small cavities initially presented in the HDPE [16,26,38,39].

3- The last stage is associated with the plastic strain hardening region, where the Poisson's ratio is a minimum close to the onset of strain hardening, then monotonically increases further in applied strain, which seems to be related to the formation and growth of cavities [26,29,37,40].

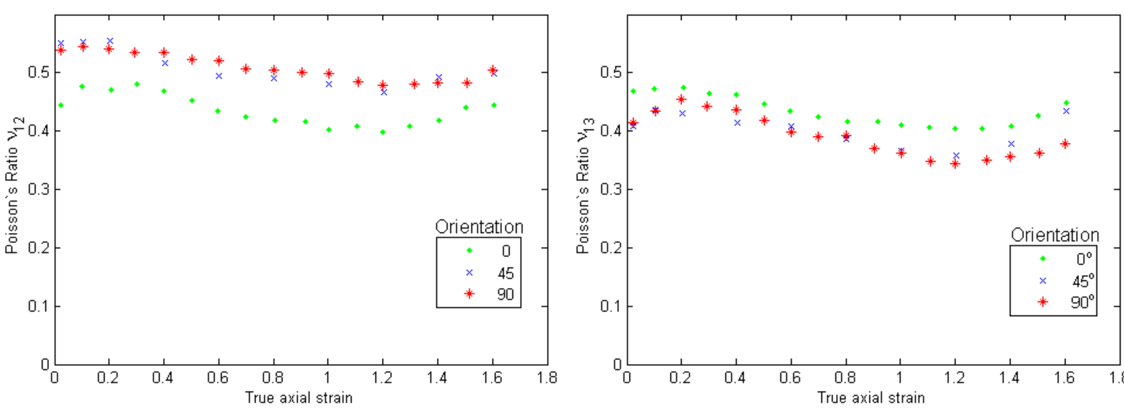

Figure 5: Poisson's ratios ν_{12} and ν_{13} for different orientations of specimens at Strain rate of $0.001s^{-1}$.

The evolutions of young's modulus for different specimen orientations are represented in Figure (6). The Young's modulus does not significantly depend on the specimen orientation, since there is a slight difference in the values of Young's modulus at different orientations. The Young's modulus measured in the present study may integrate microstructural changes occurring during the deformation, and therefore may be such a dependency exists.

Regardless of the specimen orientation, the apparent Young's modulus has similar evolution according to axial strain. Once the necking initiated, corresponds to onset of microstructural changes during plastic deformation, the apparent Young's modulus showed a remarkable decrease as axial strain increases. The apparent Young's modulus is about a fourth of its initial value for axial strain close to $\varepsilon_{11} = 0.5$, and it is independent of the anisotropy effect (orientation effects of the investigated material). This decrease of apparent Young's modulus is basically associated to structural damage that takes place during plastic deformation. Indeed, it can associate to the progressive fragmentation of crystallites and the cavitation phenomenon. This first phase is followed by a significant increase in Young's modulus until 2/3 of its initial value for large strains. The main reason for this increase is probably linked to reorientation of macromolecular chains, which become aligned with the tensile axis. Although it is not sufficient to be quantified in this study, the beginning of this augmentation is orientation dependent and can be related to strain levels known as natural draw ratio.

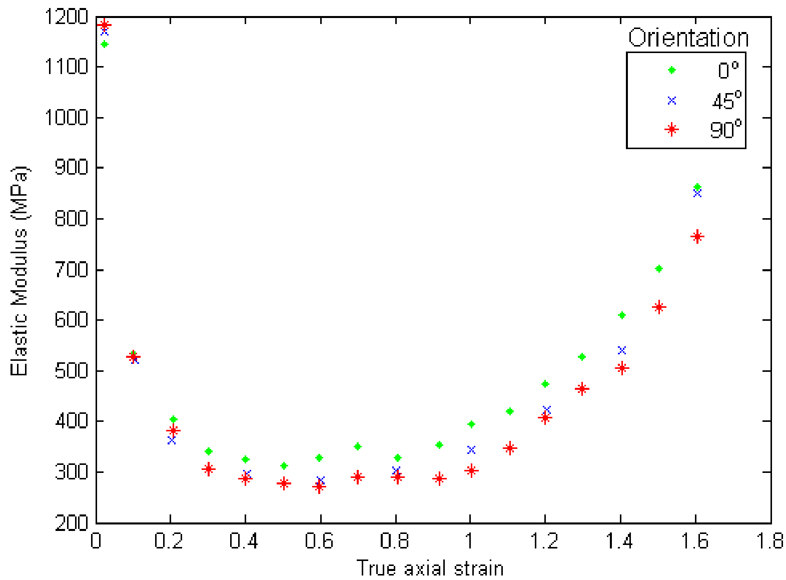

Figure 6: Evolution of elastic modulus during Step-Cycle tests for different orientations of specimens at strain rate of 0.001s^{-1}.

MODELING OF PLASTIC INSTABILITY

Like all other semicrystalline polymers, HDPE exhibits inhomogeneous plastic deformation under uniaxial tensile loading, with strain localization leading to a neck formation, stabilization and propagation. Neck initiation and neck stabilization are corresponding to the maximum and minimum of deformation force, respectively, which are generally accounted for by Considerer's criterion. This criterion is used for the prediction of plastic instability of the polymeric materials, with the assumption of constant volume for plastic deformation [7,30,31,41,42]. However, it is inefficient especially for determining the strain at minimum force, since there is some difference between the observed and calculated values due to the volume change during deformation [7,42].

The concept of the Considerer's method is based on engineering strain, while this criterion can now be developed in terms of true strain $\partial\sigma/\partial\epsilon = \sigma$, which means that the onset of macroscopic necking takes place when the instantaneous strain hardening rate, $\partial\sigma/\partial\epsilon$ is equal to the value of true stress, noting that Considère's analysis is only applicable to materials that are not sensitive to the strain rate. One of most macroscopic instability criterion to describe the necking in rate-sensitive materials such as semicrystalline polymer was developed by Hart [43] which is based on the strain hardening coefficient, $\gamma=(1/\sigma)(\partial\sigma/\partial\epsilon)_{\dot{\varepsilon}}$ and the strain rate sensitivity parameter, $m=(\dot{\varepsilon}/\sigma)(\partial\sigma/\partial\dot{\varepsilon})_{\epsilon}$ [31,44,45]. Therefore, the form of this instability criterion is:

$$\gamma + m \geq 1 \quad (1)$$

The plastic instability is considered active in the strain range when $\gamma = 1$, which typically gives a first point at the onset of the plastic instability, where the force is maximum (yield point or neck initiation), and the second point near the end of the plastic instability where the neck stabilizes, i.e., the minimum force (natural draw ratio) as shown in Figure (7). It is clear that, when the strain rate sensitivity is negligible from this criterion, equation (1) is equivalent to the classical Considère construction. Therefore, in accordance with the Considère condition for necking as proposed by Vincent [30], these maximum and minimum values are usually called 1st and 2nd Considère condition or strain respectively [41].

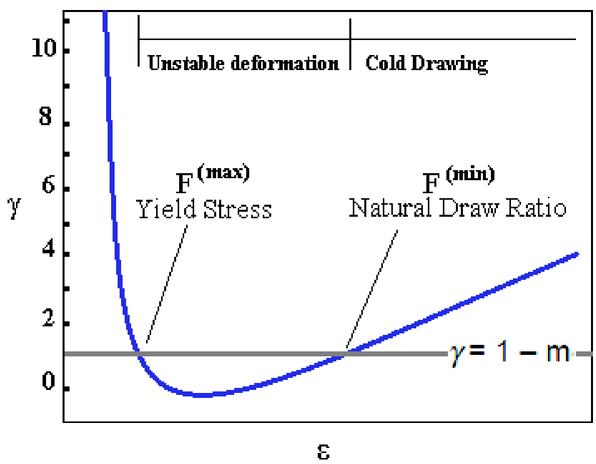

Figure 7: Strain-hardening coefficient as a function of the axial true strain, ($\gamma = 1 - m$ represents the stability criterion developed by Hart [43]).

Most criteria in the literature used for the prediction of plastic instabilities during deformation of the polymeric materials, were essentially obtained for metals. The volume variations during plastic deformation are generally neglected, which is not true for the case of semicrystalline polymers. Therefore, the aim is to revise the well known Considère criterion in order to consider the effect of volume change during plastic deformation which is one of the significant points on the way to correctly model the plastic deformation of semicrystalline polymers.

Under isothermal mechanical behavior and a constant strain rate, the onset of the instability can be defined by Considère criterion at the point where the applied force reaches a maximum and the tangent slope of the force – strain (or displacement) curve passes zero (dF= 0), then the maximal load criterion can be described as [46]:

$$dF = \sigma dA + A d\sigma = 0 \tag{2}$$

$$\frac{d\sigma}{\sigma} = -\frac{dA}{A} \tag{3}$$

Under the condition of constancy of the volume, the plastic instability will begin when the following criterion is verified:

$$\frac{d\sigma}{d\varepsilon_1} = \sigma \tag{4}$$

Experimentally, by using an optical measurement system, an accurate deformation measurement can be obtained. This allows the evaluation of volume change in a quantitative way by determining in real time the true strains in three principal directions of a small volume element located exactly where the necking initiates and propagates. The volume strain is simply computed from the trace of the true strain tensor as [8,16]:

$$\varepsilon_v = \varepsilon_1 + \varepsilon_2 + \varepsilon_3 = \ln\left(\frac{V}{V_0}\right) \tag{5}$$

Where $V_0 = A_0 L_0$ and $V = AL$ represent the initial and current volumes of the RVE, respectively. The evaluation of current area A during the tensile test is a function of principal strain components and can be written as:

$$A = A_0 (\frac{l_0}{l}) e^{\varepsilon_1 + \varepsilon_2 + \varepsilon_3} = A_0 e^{\varepsilon_2 + \varepsilon_3} \tag{6}$$

then

$$dA = A_0 e^{\varepsilon_2 + \varepsilon_3} (d\varepsilon_2 + d\varepsilon_3) \tag{7}$$

Substituting the above derivative in equation 3, the instability criteria can be expressed as:

$$\frac{d\sigma}{\sigma} = -(d\varepsilon_2 + d\varepsilon_3) \neq d\varepsilon_1 \tag{8}$$

However, the rearrangement and dividing equation 8 by $d\varepsilon_1$, respecting volume change, one can obtain the following new expression for Considère criteria as:

$$\frac{d\sigma}{d\varepsilon_1} = -\sigma \left(\frac{d\varepsilon_2}{d\varepsilon_1} + \frac{d\varepsilon_3}{d\varepsilon_1}\right) \tag{9}$$

It is obvious that, the transverse anisotropic behavior of HDPE is taken into account in the above expression by the relative dimension changes of strains in the transverse directions ε_2 and ε_3 which are given by tangential Poisson's ratio ($\nu_{t\ 12}(\varepsilon_1) = -\frac{d\varepsilon_2}{d\varepsilon_1}$, $\nu_{t\ 13}(\varepsilon_1) = -\frac{d\varepsilon_3}{d\varepsilon_1}$). The evolutions $\nu_{t\ 12}(\varepsilon_{11})$ and $\nu_{t\ 13}(\varepsilon_{11})$ can be described together by introducing mean tangential Poisson's ratio as:

$$\tilde{\nu}_t = \frac{1}{2}\left(\frac{d\varepsilon_2}{d\varepsilon_1} + \frac{d\varepsilon_3}{d\varepsilon_1}\right) \tag{10}$$

By combining the relationships 9 and 10, the modified Considère criterion considering the volume variation of polymers during the uniaxial deformation can be rewritten in the final form as:

$$\frac{d\sigma}{d\varepsilon_1} = -2\sigma\tilde{\nu} \tag{11}$$

$$\frac{d\sigma}{d\varepsilon_1} = \tilde{\sigma} \tag{12}$$

Where $\tilde{\sigma} = -2\sigma\tilde{v}$, noting that in case of constant volume where the Poisson's ratios $\nu = 0.5$, equation 9 will be reduced to the standard expression of Considère criterion $d\sigma/d\varepsilon_1 = \sigma$.

Modified Criterion and Orientation Effect

This part analyzes graphically the case of materials that exhibit volume variation during plastic deformation such as HDPE. This modified Considère criterion is expressed by the intersection of the $\tilde{\sigma}(\varepsilon)$ with $d\sigma/d\varepsilon_1$ curves in two points as shown in Figures (8, 9, 10) for different specimen orientations. The first strain ε_o is associated with the initiation of the plastic instability with maximum force, the second ε_e determines the end of the plastic instability where the force is minimum, i.e., the natural draw ratio. Note that, the difference in predictions between the present criterion and Considère criterion at constant volume is noticeable at large strain. The 1st Considère condition which appears near $\varepsilon_o = 0.1$ is the same for the two criterions, whereas for the 2nd Considère condition, the modified Considère strain precedes that of classical Considère strain.

As already mentioned above, the anisotropic effect which has a high influence on the natural draw ratio, ε_e, can be shown by this criterion. The natural draw ratio in this way is strongly dependent on orientation, since it is approximately close to 1 for angle 0° as the isotropic case, while reaches 1.27 for the anisotropic case at an angle of 90°. However, insignificant orientation dependency observed in the 1st Considère strain, as well as the onset of necking appears to be independent of orientation effect. It should be mentioned that all figures were constructed from a function-fitting procedure on the true stress-axial force versus true strain curves, using fifth-order polynomial fits.

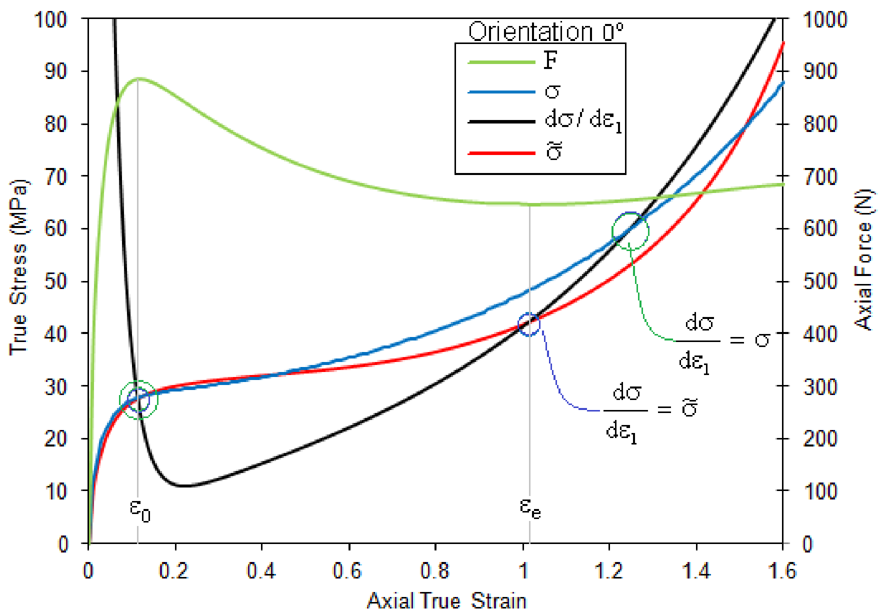

Figure 8: Determination of the onset and end of plastic instability using the modified and classical Considère criterion for angle 0° and strain rate of $0.001s^{-1}$.

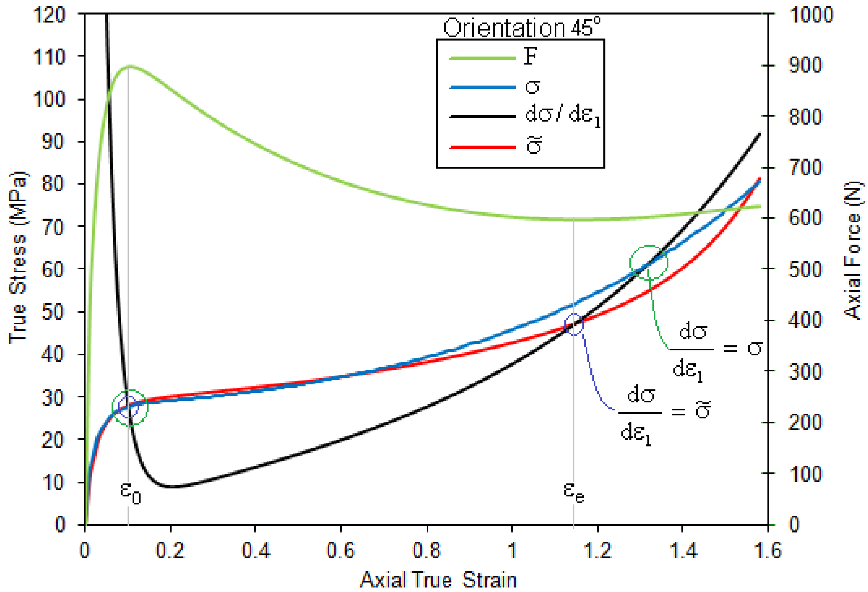

Figure 9: Determination of onset and end of plastic instability using the modified and classical Considère criterion for angle 45° and strain rate of 0.001s^{-1}.

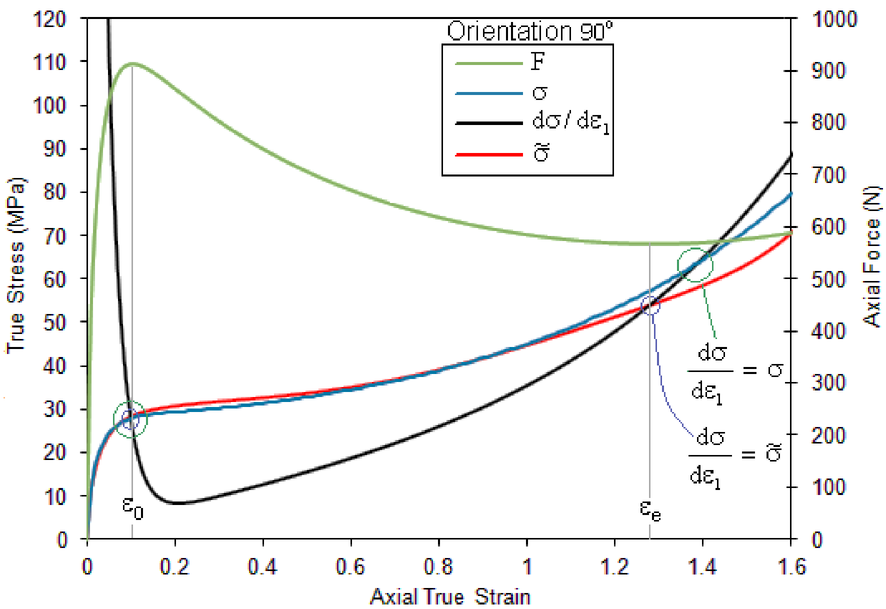

Figure 10: Determination of onset and end of plastic instability using the modified and classical Considère criterion for angle 90° and strain rate of 0.001s^{-1}.

Strain Rate Effect

True stress-True strain curves obtained from tensile testing of HDPE for three different strain rates (0.001s^{-1}, 0.005s^{-1} and 0.01s^{-1}) at controlled temperature of 23°C are shown in the Figure (11). Similar to many other polymers, HDPE has significant time dependent behavior. Also, the mechanical properties and deformation mechanisms are strongly dependent on the applied strain. There is a pronounced effect of strain rate on the true stress, since an increase in strain rate increases the true stress at all strain levels due to the decrease of molecule mobility of the polymer chains which rises the chains stiffness [23,47]. The initial young's modulus and yield stress appear also to be strain rate dependent, as shown in Figure (12). There are about 36 % increases in initial young's modulus, and about 20 % in yield stress as the strain rate increases from 0.001 s^{-1} to 0.01 s^{-1} for the two orientations. This suggests that the stiffness properties are proportional to the rate of strain, whereas the orientation of extrusion direction does not have a significant influence particularly for the initial young's modulus.

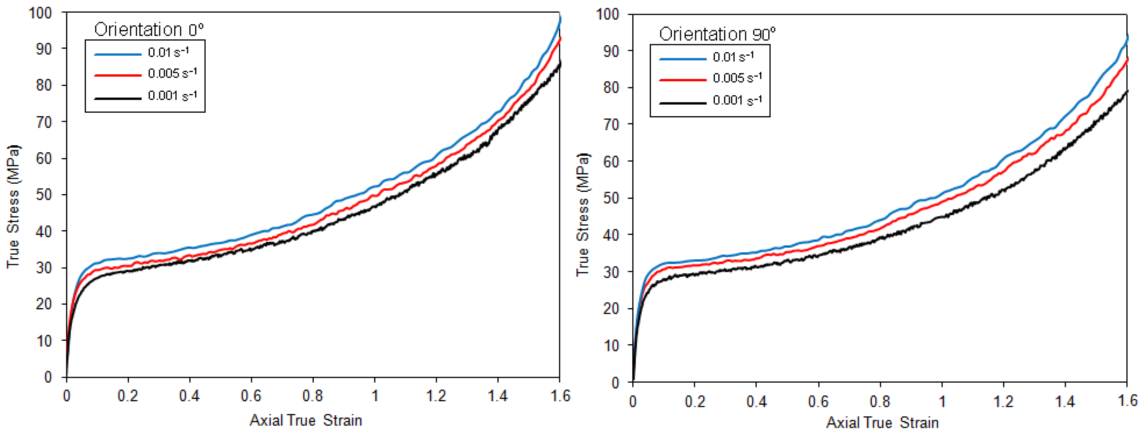

Figure 11: Influence of strain rate on mechanical behavior of HDPE at different specimen orientations.

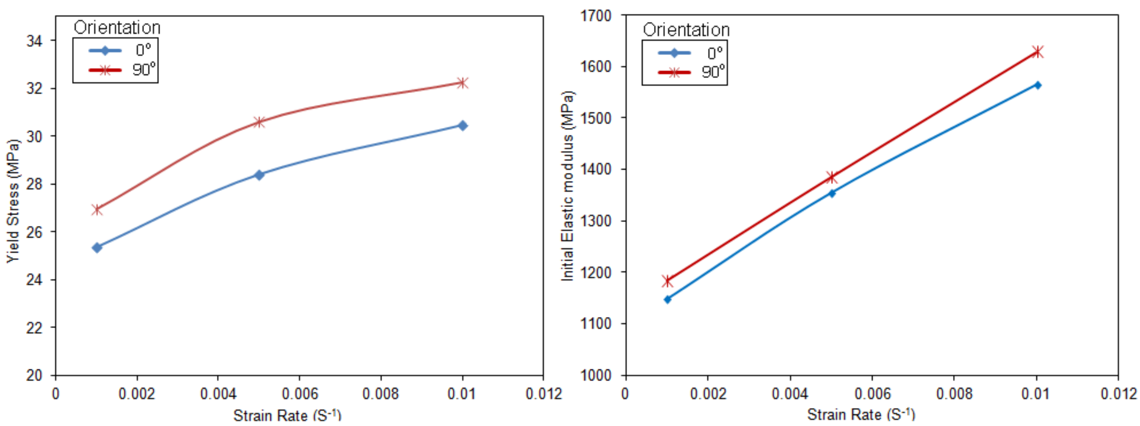

Figure 12: Influence of strain rate on a) yield stress and b) Young's modulus, at different orientations

Figure (13) illustrates the influence of strain rate on the tensile plastic instability behavior of HDPE using the modified criterion for two orientations (0˚ and 90˚). For the $\tilde{\sigma}(\varepsilon_1)$ and $d\sigma/d\varepsilon_1$ evolution, a highly dependent strain rate is observed. It can be seen that necking initiates at low strains, while at larger plastic strains the general trend becomes associated with a slight decrease in 1st Considère strain ε_o with the increase in the strain rate. An increase in the yield stress (maximum force) is also observed.

For the 2nd Considère strain, a strain rate dependency exists, and its influence on the isotropic direction (angle 0˚) is insignificant, however it has a high dependency for the anisotropic one (for angle 90˚). The onset of the phenomenon of strain hardening took place much earlier as the strain rate increases. Subsequently the natural draw ratio decreases as the strain rate increases, which indicates to a strong interior molecular chain orientation phenomenon that accompanies extension and slippage phenomena [21].

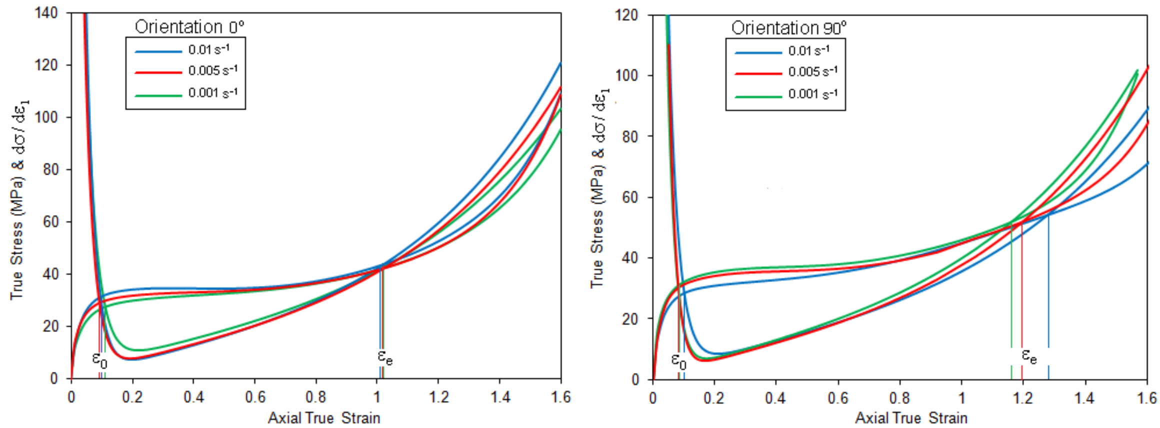

Figure 13: Influence of strain rate on the onset and end of plastic instability using a modified 'Considère criterion', for angles 0˚ and 90˚.

CONCLUSIONS

In this work, the plastic instability of high-density polyethylene HDPE was experimentally investigated. A series of uniaxial tests with several loading paths were carried out in order to determine the effect of orientation dependency and strain rates on the elasto-viscoplastic behavior during initiation and propagation of plastic instability.

The mechanical response of the HDPE is highly nonlinear and it is significantly affected by the strain rate. The experimental results showed a significant orientation dependency which is highly visible at large deformation, with no orientation effect on the onset of necking, but it was a high orientation dependency for the natural draw ratio value. The extruded HDPE showed a planar anisotropic behavior for the three orientations of specimens, and the transverse strains ε_2 and ε_3 are not equal except for the extrusion orientation (0˚) which represents the only direction of the transverse isotropy.

The apparent elastic properties of the damaged HDPE were determined during plastic instability, regardless of the specimen orientation. The Poisson's ratio showed a significant increase at low deformation and reached its maximum value at yield point, while it is significantly decreased to a minimum value during the necking initiation and propagation which is corresponding to the natural draw ratio. It was observed that the evolution of Poisson's ratio in the necking region is strongly dependent on the specimen orientation. In contrast, the slight difference in the values of the apparent Young's modulus evolution was observed at different directions.

The classical Considère criterion was revised in order to provide a modified instability criterion which is able to take into account the volume variation due to damage under large deformation of semicrystalline polymers. The results obtained for 1^{st} and 2^{nd} Considère strain based on the modified model, for various specimen orientations and strain rates, showed a good agreement with experimental data. Depending on the specimen orientation, the 1^{st} and 2^{nd} Considère strain are sensitive to the strain rate. The strain rate has a slight influence on the general trend for the isotropic direction (angle 0˚), but it has a highly dependency for anisotropic one (angle 90˚).

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