Mechanical Properties and Deformation Behaviour of Polymer Materials during Nanosectioning

Characterisation and Modelling

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Abstract

Research in local fracture processes and micro-machining of polymers and polymer-based composites has attracted increasing attention, in development of composite materials and miniaturisation of polymer components. In this thesis, sectioning (machining) of a glassy polymer and a carbon nanotube based composite at the nanoscale was performed by an instrumented ultramicrotome. The yield stresses and fracture toughness of these materials were determined by analysing the sectioning forces. Fractographic analysis by atomic force microscopy was conducted to characterise the topographies and elastic properties of the sectioned surfaces to explore the deformation and fracture behaviour of the polymer during nanosectioning. The study reveals that a transition from homogenous to shear localised deformation occurred as the uncut chip thickness (depth of cut) or sectioning speed increased to a critical value. Analytical and finite element methods were used to model the nanosectioning process. The shear localised deformation was caused by thermal softening due to plastic dissipation. Although not considering sectioning, the tensile properties of a polymer nanocomposite were additionally investigated, where the degree of nanofibrillation and polyethylene glycol (PEG) content had significant effects.

Keywords: Nanosectioning; Fracture toughness; Adiabatic shearing; Shear band; Nanosectioning; Glassy polymer; Nanocomposite

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Dedicated to my family
This thesis is based on the following papers, which are referred to in the text by their Roman numerals.


II Sun, F., Li, H., Leifer, K., Gamstedt, E.K. Rate effects on localized shear deformation during nanosectioning of an amorphous thermoplastic polymer. *International Journal of Solids and Structures*, accepted.

III Sun, F., Li, H., Leifer, K., Gamstedt, E.K. Effect of nanosectioning on surface features and stiffness of an amorphous glassy polymer. *(Submitted)*

IV Sun, F., Gamstedt, E.K. Finite element modeling of nanosectioning of a glassy polymer based on an elastic-viscoplastic model. *(Manuscript)*


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Abbreviations

AFM  Atomic Force Microscopy
CNF  Cellulose nanofiber
LEFM  Linear elastic fracture mechanics
MWCNT  Multiwall carbon nanotube
PEG  Polyethylene glycol
PMMA  Poly(methyl methacrylate)
PSZ  Primary shear zone

Symbols

$\tilde{D}_\alpha, \tilde{D}_\beta$  Plastic stretch tensor
$F, F_c, F_t$  Sectioning forces
$F_\alpha, F_\beta$  Deformation gradient
$I$  Intercept of the plot $F_c$ vs $t_u$ at $t_u = 0$
$l_s$  Length of the PSZ
$m$  Softening parameter
$n$  Molecular chain parameter
$N_\alpha^p, N_\beta^p$  Deviatoric directions
$Q$  Friction parameter
$q_0$  Heat flux
$R$  Fracture energy
$r_i$  Distance between a heat segment and point $M$
$S$  Slope of the plot $F_c$ vs $t_u$
$\tilde{s}_\alpha, \tilde{s}_\beta$  Athermal shear strength components
$T_g$  Glass transition temperature
$t_u$  Uncut chip thickness
$T_\alpha, T_\beta$  Cauchy stress tensors
$v$  Sectioning speed
$v_s$  Shear velocity on PSZ
$w_u$  Width of cut
$Z$  Dimensionless parameter for sectioning
$\alpha$  Rake angle of the knife
\( \alpha_{p,\alpha}, \alpha_{p,\beta} \)  \quad \text{Hydrostatic pressure coefficients} \\
\( \alpha_i \)  \quad \text{Thermal diffusivity} \\
\( \beta \)  \quad \text{Coulomb friction angle} \\
\( \gamma \)  \quad \text{Plastic strain} \\
\( \dot{\gamma}_{0,\alpha}, \dot{\gamma}_{0,\beta} \)  \quad \text{Pre-exponential constants for plastic strain rate} \\
\( \Delta G_\alpha, \Delta G_\beta, \Delta H_\beta \)  \quad \text{Activation energy of deformation} \\
\( \varepsilon \)  \quad \text{Strain} \\
\( \dot{\varepsilon} \)  \quad \text{Strain rate} \\
\( \dot{\varepsilon}_0 \)  \quad \text{Pre-exponential constant for strain rate} \\
\( \theta_M \)  \quad \text{Temperature rise at point M} \\
\( \lambda \)  \quad \text{Thermal conductivity} \\
\( \sigma \)  \quad \text{Stress} \\
\( \sigma_1, \sigma_2 \)  \quad \text{Principle stress components} \\
\( \sigma_i(0) \)  \quad \text{Athermal yield stress} \\
\( \sigma_m \)  \quad \text{von Mises stress} \\
\( \sigma_y \)  \quad \text{Yield stress} \\
\( \tau_{\text{Bulk}} \)  \quad \text{Shear stress in the bulk material} \\
\( \tau_{\text{PSZ}} \)  \quad \text{Shear stress in the primary shear zone} \\
\( \tau_y \)  \quad \text{Shear yield stress} \\
\( \phi \)  \quad \text{Shear plane angle} \\
\( \phi' \)  \quad \text{Second shear plane angle}
1 Introduction

Amorphous glassy polymers such as polycarbonate (PC) and poly(methyl methacrylate) (PMMA) possess low densities, high transparency and excellent mechanical properties [1], and are extensively used in commercial and military applications, ranging from vehicle windshields, eyeglasses to intraocular lenses and body armour. Although the mechanical properties of amorphous polymers have been investigated intensively, the issue of the measurement of the fracture toughness at a microscopic level is still not well solved yet. Conventional linear elastic fracture mechanics (LEFM) for macroscopic testing the fracture toughness was initially established for metals in load carrying structures, and then introduced to polymers and other materials. The fracture toughness of ductile materials measured by LEFM-based methods is two to three orders of magnitude greater than the theoretical surface free energies (a few J/m²) [2]. Orowan and Irwin argued that the discrepancy between the experimental and theoretical values of surface energy is due to the large plastic deformation near the crack surface [3]. LEFM testing of polymers is always accompanied with the problem of crack blunting with large-scale yielding and craze occur at the crack tip [3] (see Fig. 1), which arises an overestimation of the toughness. For toughened polymers and polymer blends other approaches e.g. the J-integral, essential work of fracture and crack opening displacement have been developed.

**Figure 1.** Schematic diagram of process zone in ductile fracture specimen (Source: J. Wu and Y.-W. Mai [4]).
More recently, the sectioning (cutting) method is becoming an alternative approach for the measurement of fracture toughness for polymers. Sectioning is a controlled material-removal process, separating a layer of material from the bulk in forms of chips, which can be viewed as a crack propagation process. This method can avoid the problem of large-scale yielding in front of the crack tip since the knife edge can touch the crack tip during the whole process [5]. Previously, Ericsson and Lindberg [6] used a nanosectioning method to investigate the energy dissipation mechanisms in polymers. Atkins [2,7] reformulated the energy balance equation by taking fracture dissipation into account in the sectioning analysis, and concluded that the fracture toughness of material can be determined from the sectioning force. More recently, Williams and his co-workers [4,8,9] improved this methodology. To date, the sectioning method has been applied to determine fracture toughness of metals [2,10], polymers [5,11], nanocomposites [12,13], wood [14], body tissues and bones [15,16], etc.

The fast development in device miniaturisation demands increased abilities to manipulate matter at the nanoscale and even the atomic level. Sectioning of polymers at the nanoscale (sub-microscale) is of great significance in manufacturing components and devices for electrical and optical applications [17]. As the sectioning scale goes down to the nanoscale, the terms related to the volume (e.g. the plastic work) are strongly restricted while the terms related to surface area (e.g. fracture) becomes relatively more pronounced, and special deformation behaviour is anticipated to occur.

1.1 Plastic behaviour of amorphous thermoplastics

In this section, the plastic behaviour of amorphous thermoplastic polymers is recapitulated. Polymers can deform plastically, with chain molecules sliding past each other over large distances. As glassy polymers undergo large deformation, two kinds of physical resistance must be overcome before large inelastic flow occurs. Below the glass transition temperature \( T_g \), prior to initial yield, the material needs to be stressed to exceed its intermolecular resistance to segment rotation. Once the material is free to flow, molecular alignment occurs, resulting in an anisotropic internal resistance to further inelastic deformation [18].

The plastic behaviour of amorphous polymers strongly depends on the temperature because obstacles have to be overcome by thermal activation. Far below \( T_g \), chain molecules cannot easily slide past each other because the bonds between molecules are very strong and the specific volume (the reciprocal of the density) for movement is too small. Under loading, brittle failure usually takes place by breaking the intermolecular bonds (Fig. 2a). When the temperature approaches about 0.8\( T_g \), the molecules gain some mobility and
the polymers exhibit limited ductility (Fig. 2a). Glassy polymers can deform by shear banding or crazing. Formation of shear bands is especially important under compressive loads, while craze can only form under hydrostatic tensile stress [19], as illustrated in Fig. 2b. When the temperature approaches \(T_g\), the chain molecules become more mobile and can rearrange on loading. Under sufficient large deformation, molecular chains are drawn and orientated in parallel, leading to local hardening. If the temperature exceeds \(T_g\), the molecules obtain very high mobility and polymers behave almost like viscous liquids.

Figure 2. Mechanical behaviour of glassy polymers (a) at temperature far below \(T_g\) and around 80% of \(T_g\), and (b) yield surface of glassy polymers that can fail by crazing or shear banding, in which \(\sigma_m\) is the hydrostatic pressure. (after Roesler et al. [19])

Varying the strain rate also influences the deformation mode. Increasing the testing time i.e. decreasing the strain rate is equivalent to increase the temperature. Taking the tensile test of a glassy polymer sample at the room temperature for example, the sample is ductile and cold-draws at low strain rates, whereas it exhibits brittle fracture behaviour at high rates. Changing the strain rate may lead to the isothermal–adiabatic transition. Under plastic deformation with high strain rates, the heat converted from the plastic work cannot conduct to the surrounding material rapidly due to the low thermal diffusivities (~\(10^{-7}\) m\(^2\)/s, two orders of magnitude lower than those in metals [20]), and thus thermal softening results (illustrated by the compressive behaviour of PMMA at strain rate above 1000 /s in Fig. 3), which leads to ductile or even localised deformation [21].

In addition, a glassy polymer exhibits apparent scale effects (or cube-square scaling effects) [22], namely it tends to deform in a ductile manner at small scale volume while behaves in a brittle fashion at large scale. Since sectioning is a shear dominated dynamic process, the glassy polymer may exhibit different deformation modes when the sectioning speed or uncut chip thickness (depth of cut) varies, which directly influence the surface qualities in manufacturing of engineering applications.
1.2 Sectioning

1.2.1 Chip types

The chips formed by sectioning can be broadly classified into four types [24], as illustrated in Fig. 4.

(a) Continuous chip: this type of chip is usually produced in sectioning of ductile materials, indicating a steady sectioning process and yielding good quality surfaces.

(b) Continuous chip with built-up edge: these chips are formed in ductile, work-hardening materials at low sectioning speeds, with parts of the work material welding on the tool edge and becoming a part of the tool tip.

(c) Discontinuous chip: discontinuous chips are frequently formed in the sectioning of brittle materials at low speeds.

(d) Shear localised chip: this type of chips are macroscopically continuous, consisting of narrow bands of heavily deformed material.
These chips can be obtained in sectioning hardened and stainless steels and titanium alloys at high speeds.

Figure 4. Types of chips formed during sectioning of metals, (a) continuous chip, (b) continuous chip with built-up edge, (c) discontinuous chip, and (d) shear localised chip.

1.2.2 Overview of analysis methods

In the past few decades, sectioning mechanisms have been investigated intensively and many important models have been developed. Pisspanen [25] proposed the ‘pack of cards’ model to depict the shear deformation during sectioning process. Merchant [26] developed the well-known force circle and derived the shear plane angle by adapting the principle of minimum energy. Both of the ‘pack of cards’ model and Merchant’s force circle are viewed as cornerstones for sectioning analysis. In the 1950s, Lee and Shaffer introduced the slip-line field theory to sectioning analysis, which brings the analysis to 2D modelling. Since then, many efforts have been made to construct more complex and accurate slip-line models for sectioning [27–30], one of which is the Oxley’s parallel shear zone model accounting for strain hardening mechanisms [31]. A more universal slip-line model for sectioning analysis was developed by Fang et al. [32] in 2001.
With regard to the formation of shear localised metal chips, Recht [21] proposed an adiabatic heating induced shear instability model. Based on Recht’s hypothesis, Komanduri and Hou [33,34] developed a classical adiabatic shearing model to predict the onset of shear localised chips. Another model, shear cracking model, was proposed by Nakayama [35], based on experimental work on highly cold worked (brittle) brass machined at very low speeds. This model was later improved by Shaw [36]. The essential distinction between the two theories lies in the root cause of the shear band. While the shear band initiates from the knife-tip in the adiabatic shearing theory, it originates from the free surface of the chip in the shear cracking theory. Apart from these two models, Sullivan et al. [37] developed the slip-stick friction model, Davies and Burns [38,39] proposed the loading-unloading (reactions between the tool surface and material) model to describe the formation of shear localisations, etc.

With the fast development of the computer technology, finite element methods have been widely used in analysing the sectioning process since 1980s [24,40–42]. Finite element simulation is now viewed as a reliable approach to analyse the sectioning process. With finite element analysis, the complex large elastic-plastic deformation, contact/friction, thermo-mechanical coupling and chip separation mechanisms, which occur in the vicinity of the cutting edge and are not directly observable, can be better described and understood [24]. To date, the Eulerian method, Lagrangian method and coupled Eulerian-Lagrangian method, smoothed particle hydrodynamics, etc. have been used in the analyse of the sectioning process [43,44]. Recently, researchers have attempted to use molecular dynamics methods to investigate the sectioning conducted at the nanoscale [45,46].

1.3 Objective

Investigating the mechanical response of glassy polymers during sectioning process is important for both scientific studies and engineering applications. The main objective of this thesis is to reveal the mechanisms that govern the deformation and fracture behaviour of polymeric material during nanosectioning by experiment and modelling. Such knowledge can be useful in controlling and limiting damage formation in manufacturing of small-scale polymer components, and in models predicting the mechanical behaviour on the sub-micrometre level in polymer-matrix composites.

1.3.1 Assessing fracture properties of a glassy polymer

In this subtopic, the fracture toughness and energy dissipation mechanisms during sectioning at the nanoscale are investigated. The main work
includes: (1) the design and development of robust apparatus for performing nanosectioning and measurements of sectioning forces, chip thicknesses, etc.; (2) assessing the fracture toughness of the glassy polymer, and (3) characterising the deformation and fracture behaviour of the glassy polymer during nanosectioning.

1.3.2 Rate dependence of sectioning

As be described in Section 1.1, the mechanical behaviour of glassy polymers is very sensitive to the strain rate. In this subtopic, the influence of strain rate on the material deformation behaviour during sectioning process is investigated by conducting nanosectioning with varying speed. Modelling work is also carried out to explore the possible mechanisms that control the material response.

1.3.3 Influence of nanosectioning on surface properties

Acquiring high-quality surfaces (good integrity, high mechanical and optical properties) by sectioning is a main concern for engineering applications. In this subtopic, the influence of the sectioning condition (uncut chip thickness, sectioning speed) on the surface elasticity and damage of a glassy polymer is investigated by experiment and modelling.

1.3.4 Finite element modelling of polymer sectioning

Using finite element methods to investigate polymer sectioning process is important because sectioning is a complex and nonlinear process which is difficult to observe by experiment. One aim of this study is to implement finite element analysis of sectioning of glassy polymer using appropriate models including the effects of strain rates, temperature, hydrostatic pressure, etc. Numerical modelling is challenging since the mechanisms in sectioning are complex and interacting. As a first step, it is useful to see if the numerical simulations can recreate the experimentally observed phenomena. Once the model has been validated, it may be used to predict formation of damage based on material properties and sectioning conditions.

1.3.5 Testing nanocomposites by sectioning

To investigate the strengthening and toughening mechanisms of nano-fillers in composites, nanosectioning is performed on multiwall carbon nanotube (MWCNT) based composites. The focus of the study is to characterise the mechanical properties of the nanocomposites and to explore the role of MWCNT in enhancing the material. The idea is to illustrate how the nano-
sectioning method can be used to evaluate more complex materials than neat amorphous thermoplastics, e.g. composite materials.

1.3.6 Optimising cellulose nanofibre biocomposites

Wood nanocellulose has been proposed for wound dressing applications because of its capability to form translucent films and aerogels with good liquid absorption capabilities. Understanding the mechanical properties of nanocellulose films are most important for tailoring optimizing wound dressing structures with adequate strength, conformability, porosity and exudate management. Mechanical properties are usually assessed in standard conditions (50% relative humidity), which is not relevant in a wound management situation. In this study, the effect of nanofibrillation and of polyethylene glycol (PEG) addition on the mechanical properties of nanocellulose films are assessed in wet conditions.

Although nanosectioning was not used in this study, this piece of work is included in the thesis to show an example of an application of a polymer-based material, whose functionality is controlled by damage and deformation processes on the sub-micrometre scale. A nanocomposite is a heterogeneous material, where the reinforcement brings about high local stresses and eventually fracture and damage processes at a very local level.
2 Materials and methods

2.1 Nanosectioning setup

Fig. 5 displays the instrumented ultramicrotome built for the evaluation of nanosectioning. It was developed based on the previous work of Ericson and Lindberg [6], and it has been improved by us to provide more functions. On the instrumented ultramicrotome, a pair of piezoelectric force sensors (PCB 209A12) were installed in parallel in a sample holder to measure the sectioning forces, as shown in Fig. 5b. A CCD eyepiece camera (ToupTek, S3CMOS) was installed on the microscope of the ultramicrotome to provide a quantitative measure of the sizes of the chips. Other accessories for force measurement include signal amplifiers (PCB 480E09), data acquisition device (Agilent U2352A). The instrumented ultramicrotome can implement the nanosectioning by advancing the knife from 5 to 200 nm per stroke using a precisely thermal feed control.

![Experimental setup for nanosectioning, (a) an ultramicrotome instrumented with piezoelectric sensors, and (b) the sample holder.](image-url)
When the knife sections the material, the resultant force $F$ acting on the work material can be resolved into two orthogonal components: the force component parallel to the sectioning direction $F_c$ and the one normal to the sectioning direction $F_t$. In each sectioning, one pair of voltage signals, $V_1$ and $V_2$, was obtained from the force sensors, as illustrated in Fig. 6a. The force $F_c$ and $F_t$ were proportional to the absolute values of $V_1-V_2$ and $V_1+V_2$, respectively. This relation was determined through the sensor calibration with known weights (see Fig. 6b), showing a force resolution down to 1 mN.

**Figure 6.** The signal acquisition and calibration, (a) output signals during sectioning, and (b) the relationship between the voltage signal and the load.

*Atomic Force Microscopy* After nanosectioning, the sectioned surfaces of the work material and the back sides of chips (in contact with the knife) were examined by AFM (Multimode 8, Bruker) in the ScanAsyst mode based on a peak force tapping mode, using a silicon tip with a radius of 3 nm.

*Nanoindentation* To verify the shear stress determined by nanosectioning, the nanoindentation testing was applied to characterise the mechanical properties of material. Nanoindentation was performed using a CSM UNHT instrument equipped with a 40 nm diamond cube corner tip. A strategy of quick loading and unloading with no holding segment was deliberately employed to minimize viscoelastic relaxation during testing to push the nanoindentation towards the quickest process possible, in an attempt to mimic the sectioning process. Ten indents were made in each composite, all to a depth of 150 nm and with a loading and unloading rate of 1 μm/min.

### 2.2 Materials

#### 2.2.1 PMMA

A commercial Quinn XT extruded PMMA sheet (2 mm thickness) manufactured by Quinn Plastics was chosen to perform the nanosectioning. The
PMMA samples were prepared in the following procedure. Adjacent cubic blocks with a size of 5×5 mm were cut out from the sheet. A protruded mesa was trimmed out at the front centre of each block using pristine glass knives by ultramicrotomy, as shown in Fig. 7.

![Figure 7. Prepared sample for nanosectioning, (a) a PMMA block, and (b) the mesa on the PMMA block.](image)

2.2.2 MWCNT/PMMA

In this study, MWCNT/PMMA composites produced by CICY (Mexico) were chosen to perform the nanosectioning. The MWCNT content of the composites varied from 0.25 to 1.0 wt%. The raw materials that were used in the manufacturing includes: (1) commercial multi-walled carbon nanotubes (MWCNTs) (Cheaptubes Inc., USA) produced by chemical vapour deposition with 30-35 nm outer diameter, 5-10 nm inner diameter, 1-6 μm length and purity > 95%; (2) commercial H15 002 PMMA acquired from Plastiglas (Mexico, DF); (3) the standard reagent of Chloroform with 98.8% purity.

2.2.3 Cellulose nanofibre composites

The mechanical properties cellulose nanofibre (CNF) composites were also investigated. The CNF was treated with different number of passes in the homogenisation process. Two groups of CNF films are produced. In one group, the films were made of pure CNFs, and in the other group 40 wt% polyethylene glycol (PEG) were added.

2.3 Analysis methods

(1) Atkins’ model (more details can be found in [2]) was used to determine the fracture toughness (specific work of surface formation) and shear yield stress of material. (2) An adiabatic shearing model developed by Komanduri and Hou (see [33]) was adopted to analyse the plastic properties of
PMMA under adiabatic heating with a constitutive model proposed by Richeton et al. [47]. (3) A damage model was used in analysing the effects of sectioning speed on the surface elasticity of PMMA.

Finite element analysis was conducted to explore the deformation behaviour of PMMA during sectioning, in which the experimental stress-strain plots obtained by Richeton et al. [23] and a modified Mulliken-Boyce model (see [48]) were used to describe the yield and post-yield response of PMMA under high strain rates.

2.3.1 Atkins’ model

The external work during sectioning is considered to dissipate by the plastic deformation in the shear zone, friction on the chip-knife interface and crack propagation ahead of the knife (Fig. 8).

\[
\frac{F_v}{\text{Ext. power}} = \left( \tau_y \gamma (t_u w_u v) + [F_c \sec(\beta - \alpha) \sin \beta] \frac{v \sin \phi}{\cos(\phi - \alpha)} \right) + \frac{R w_u v}{\text{Fracture}}
\]

where \(\tau_y\) is the shear yield stress, \(\gamma\) is the plastic strain, \(R\) is the fracture energy or specific work of surface formation (work divided by area of fracture surface), \(\beta\) is the Coulomb friction angle, \(\alpha\) is the rake angle of the knife, \(\phi\) is the shear plane angle, \(t_u\) is the uncut chip thickness, \(w_u\) is the width of cut and \(v\) is the sectioning speed. Eq. (6) can be furtherly written as
where \( Q = [1 – \sin(\beta)\sin(\phi)\cos(\beta – \alpha)\cos(\phi – \alpha)] \) is a friction parameter. Williams et al. [49] derived the closed-form solution for \( \phi \) as follows

\[
cot(\phi) = \tan(\beta – \alpha) + \sqrt{1 + \tan^2(\beta – \alpha) + Z[\tan(\beta – \alpha) + \tan(\alpha)]}
\]

where \( Z = R/\tau_y t_u \) is a dimensionless parameter. The values of \( \tau_y \) and \( R \) can be obtained accordingly when the value of \( \phi \) is determined. The calculation procedure is briefly described in Fig. 9 (\( I \) and \( S \) are the intercept and slope of the \( F_c \) vs. \( t_u \) plot, respectively).

**Figure.** 9. The calculation procedure for the determination of shear yield stress and fracture toughness by sectioning.

### 2.3.2 Adiabatic shearing model

During plastic deformation, temperature rise plays a negative effect on the material strength and if it overweighs the positive effect of strain rate hardening, the catastrophic shear instability occurs [21]. Therefore, if the
stress in the primary shear zone (PSZ), $\tau_{PSZ}$, is surpassed by the stress in the bulk material, $\tau_{Bulk}$, a shear localisation event takes place.

The effects of sectioning speeds on temperature rise in the PSZ and in the bulk ahead were analysed using Komanduri-Hou model [34]. The model divides the formation of shear band into the shearing stage (Fig. 10a) and flattening stage (Fig. 10b), and each includes two primary heating mechanisms. In the shearing stage, the plastic deformation in the PSZ (S1) and the friction on the interface between the chip segments already formed and the knife (S3) are the main heating mechanisms. In the flattening stage, the shearing on the second shear plane (S2) and the friction on the interface of the chip segment being formed and the knife (S4) are the other two primary heat sources. An imaginary part of the S1 heat source which is close to an adiabatic boundary is also included (Fig. 10c). For instance, the temperature rise $\theta_M$ at $M(x, y)$ due to the first source is formulated in an integral form,

$$
\theta_M = \frac{q_0}{2\pi\lambda} \left(1 - \frac{t}{2t_0}\right) \int_{y_i=0}^{y_i'} \Omega(p)dy_i + \frac{q_0}{16\pi\lambda\alpha t_0} \int_{y_i=0}^{y_i'} r_i^2 \chi(p)dy_i
$$

(2.1)

where $q_0$ is the heat flux, $\lambda$ is the thermal conductivity, $t$ is the duration of the heat source, $t_0 = l_s/v_s$ ($l_s$ is the length of the PSZ, $v_s$ is the shear velocity on PSZ), $\alpha$ is the thermal diffusivity, and

$$
\Omega(p) = \int_p^\infty \exp(-u^2) / u^2 du, \quad \chi(p) = \int_p^\infty \exp(-u^2) / u^3 du
$$

in which $p = r_i/\sqrt{4\alpha t}$ is a non-dimensional value and $r_i = \sqrt{x^2 + (y - y_i)^2}$ is the distance between the heat segment and the point $M(x,y)$.

Figure 10. Schematics of the shear banding during sectioning, (a) the shearing stage, (b) the flattening stage, and (c) coordinate for S1 heat source. (After Hou and Komanduri [34])

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The yield stress of PMMA, $\sigma_y$, below the glass transition temperature is expressed as [47]

$$\sigma_y = \sigma_i(0) - mT + \frac{2k_BT}{V_a}\sinh^{-1}\left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0 \exp(-\Delta H_\beta/k_BT)}\right)^{1/n} \quad (2.2)$$

where $\sigma_i(0)$ is the athermal yield stress, $m$ is a softening parameter, $k_B$ is the Boltzmann constant, $V_a$ is the activation volume, $\dot{\varepsilon}$ is the strain rate, $\dot{\varepsilon}_0$ is a pre-exponential constant, $\Delta H_\beta$ is the activation energy and $n$ is a molecular chain parameter.

2.3.3 Modified Mulliken-Boyce model

The one dimensional rheological interpretation of Mulliken-Boyce model [50] is illustrated in Fig. 11. In this model, Phase A can be decomposed into two components, $A_\alpha$ and $A_\beta$, acting in parallel to describe the elastic-viscoplastic response of polymers. Both $\alpha$ and $\beta$ components are represented by a linear spring with a dashpot acting in series. Phase B i.e. a Langevin spring is used to describe the entropic-hardening process.

**Figure 11.** Mulliken-Boyce model for the description of rate-dependent elastic viscoplastic behaviour in amorphous glassy polymers.

The deformation gradient $F$ for the $\alpha$ and $\beta$ components are decomposed into elastic and plastic parts as,

$$F_\alpha = F^e_\alpha F^p_\alpha$$
$$F_\beta = F^e_\beta F^p_\beta$$

The plastic stretch is defined as
\[ \mathbf{D}^p = \dot{\gamma}^p \mathbf{N}^p \]
\[ \mathbf{D}^p = \dot{\gamma}^p \mathbf{N}^p \]

(3.2)

where \( \mathbf{N}^p_\alpha \) and \( \mathbf{N}^p_\beta \) are taken to be coaxial with the deviatoric stresses acting on the \( \alpha \) and \( \beta \) components, respectively

\[ \mathbf{N}^p_\alpha = \frac{T^\alpha_\Lambda}{T^\Lambda_\alpha} \]
\[ \mathbf{N}^p_\beta = \frac{T^\beta_\Lambda}{T^\Lambda_\beta} \]

(3.3)

Varghese and Batra [48] modified the flow rule by including new internal variables to characterise the viscoplastic behaviour,

\[ \dot{\gamma}^p_\alpha = \dot{\gamma}^p_0,\alpha \exp \left[ -\frac{\Delta G_\alpha}{k\theta} \left( 1 - \frac{\tau_\alpha}{t_\alpha \hat{s}_\alpha + \alpha_p,\alpha P} \right) \right] \]
\[ \dot{\gamma}^p_\beta = \dot{\gamma}^p_0,\beta \exp \left[ -\frac{\Delta G_\beta}{k\theta} \left( 1 - \frac{\tau_\beta}{t_\beta \hat{s}_\beta + \alpha_p,\beta P} \right) \right] \]

(3.4)

where \( \dot{\gamma}^p_0,\alpha \) and \( \dot{\gamma}^p_0,\beta \) are the pre-exponential factors, \( \Delta G_\alpha \) and \( \Delta G_\beta \) are the activation energies, \( P \) is the pressure, and \( \alpha_p,\alpha \) and \( \alpha_p,\beta \) are the hydrostatic pressure coefficients. The symbols \( \hat{s}_\alpha \) and \( \hat{s}_\beta \) denote athermal shear strengths, and evolve as

\[ \hat{s}_\alpha = \frac{0.077 \mu_\alpha}{1 - \nu_\alpha} \]  
\[ \dot{s}_\alpha = \frac{h_\alpha}{s_\alpha} \left( 1 - \frac{t_\alpha}{t_{ss,\alpha}} \right) \dot{\gamma}^p_\alpha \]

(3.5a)

\[ \hat{s}_\beta = \frac{0.077 \mu_\beta}{1 - \nu_\beta} \]  
\[ \dot{s}_\beta = \frac{h_\beta}{s_\beta} \left( 1 - \frac{t_\beta}{t_{ss,\beta}} \right) \dot{\gamma}^p_\beta \]

(3.5b)

Taking the dissipated plasticity into account and assuming the dissipation as an adiabatic process, the temperature evolution is governed by

\[ \dot{T} = \frac{1}{\rho C} \left[ \text{tr}(\mathbf{T}_\alpha \mathbf{D}^p_\alpha) + \text{tr}(\mathbf{T}_\beta \mathbf{D}^p_\beta) \right] \]

(3.6)
where $c$ is the specific heat, $\rho = \rho_0/\text{det}(F)$ is the material density in the current configuration, $T_\alpha$ and $T_\beta$ are the Cauchy stresses in each component.
3 Results and discussions

3.1 Mechanical properties of PMMA (Paper I)

The fracture toughness $R$ and shear yield stress $\tau_y$ of PMMA under the nanoscale deformation were determined by nanosectioning test. By analysing the measured sectioning forces (Fig. 12) using Atkins’ model [2], the values of $R$ around 6.4 J/m$^2$ and $\tau_y$ of 110-114 MPa were obtained for the present PMMA. It is notable that the value of $R$ in nanosectioning and the theoretical surface free energy (~1.5 J/m$^2$) are in the same order of magnitude [3].

![Figure 12. Force components at different uncut chip thickness, in group 1 the sectioning was performed on one PMMA sample and in group 2 was on individual samples.](image)

A transition of the surface feature was found at a critical uncut chip thickness. Fig. 13a displays the morphology of the surface sectioned at 60 nm under AFM, which is flat and smooth. As the uncut chip thickness increased to 85 nm, short and weak wave-like features began to appear on the surface. As the thickness increased beyond 110 nm, pronounced periodic features formed on the surfaces, as illustrated in Fig. 13b. These features oriented in parallel and were perpendicular to the sectioning direction. The average spacing between adjacent features exhibited a linear dependence on the uncut chip thickness. The features were similar to these in metals and polymer composites [12,51,52], which were attributed to the adiabatic shearing on the primary shear plane.
3.2 Rate effect on shear banding (Paper II)

Nanosectioning was performed on PMMA with the speed varying from 0.25-10 mm/s, at an uncut chip thickness of 85 nm. Features of shear band were observed to form on the sectioned surface at the speed of 1.0 mm/s, below which no bands formed (Fig. 14a). Analytical modelling of the onset of shear bands was conducted using Komanduri and Hou’s adiabatic shearing model [33]. The rate and temperature dependent yield stress of PMMA was depicted by a constitutive model proposed by Richeton et al. [23].

The temperature and yield stress in the primary shear zone (PSZ) and in the bulk ahead of the PSZ were calculated separately. Predictions showed that the onset speed for shear banding in PMMA sectioning was 4-5 mm/s, above which the yield stress in the PSZ was exceeded by that in the bulk (see Fig. 14b), and the plastic instability would take place. The modelling result agrees with the experimental result of ~1 mm/s as indicated in Fig. 12a.

![Figure 14. Shear banding in PMMA sectioning, (a) experimental surface heights along the sectioning direction, (b) the shear stresses in the PSZ and in the bulk material as a function of the sectioning speed.](image-url)
3.3 Surface properties after sectioning (Paper III)

The effects of (i) uncut chip thickness and (ii) sectioning speed on the surface stiffness of PMMA after sectioning were investigated. The effective elastic moduli of surfaces sectioned at the thicknesses varying from 85 to 200 nm and at the speed varying from 0.25 to 10.0 mm s$^{-1}$ were measured by AFM. Finite element simulation was conducted to investigate the sectioning process, and the effect of sectioning speed on the surface elasticity was analytically modelled using damage theory.

Fig. 15a shows the average effective elastic moduli of the sectioned surfaces created with varying thickness. A transition in the elastic modulus occurs at the thickness of 140 nm. Proximately below the thickness of 140 nm, the modulus decreased as the thickness increases, while above this thickness, the modulus increased. The mean elastic moduli of sectioned surfaces with varying sectioning speeds are displayed in Fig. 15b. The modulus decreased from ~3 to 2.5 GPa as the speed was increased from 0.25 to 10.0 mm/s.

![Figure 15](image)

**Figure 15.** Elastic moduli of surfaces created (a) at varying thicknesses, and (b) at varying speeds.

Abaqus/Explicit was used to simulate the sectioning process of PMMA. In lack of experimentally characterised constitutive relations of the present PMMA quality, the experimental strain-stress relationship obtained by Richeton et al. [23] was adopted for the description of the yield and post-yield behaviour of PMMA (see Fig. 3). The simulated strain distribution of the sectioning is shown in Fig. 16. As shown in Fig. 16a, localised deformation is taking place, which initiates from the knife-tip and propagates to the free surface of PMMA. With the knife advancing, the width of localised band increases, with intense plastic strain developed in the band in Fig. 16b. During the formation of a localised band, a crack pattern is formed on the sectioned surface simultaneously [53,54]. The simulations reveal the formation of localised shear bands in the chip and of similar periodic patterns on the created surface during sectioning of PMMA, which was also observed in the practical experiments (e.g. Fig. 13b).
3.4 FE modelling of shear banding (Paper IV)

In this part, the modified Mulliken-Boyce model [48,50] was used to describe the elastic-viscoplastic response of PMMA including the heat generated due to plastic dissipation under high strain rate deformation. A user defined subroutine (VUMAT) was written to implement the modified Mulliken-Boyce model by Abaqus/Explicit. An equivalent strain based fracture criterion was applied to the predefined sectioning plane where the failed elements were removed prior to the next step of calculation.

Contours of maximum strain from the sectioning at the uncut chip thickness of 85 nm and sectioning speed of 10 mm/s are shown in Fig. 17. As the knife advances, material ahead of the knife-tip is separated and flows along the rake face of the knife. Local intense strain develops near the knife-tip, and propagates towards the free surface of the work material at a certain angle relative to the cutting plane. With knife advancing, a shear band takes shape. The present FE simulation of sectioning process is capable of capturing the formation of shear bands which have been observed in our previous experiments.

The geometric characteristics of the shear band can be quantified from the simulation results. Fig. 17a shows that the shear band develops at an inclination angle of ~55° relative to the sectioning direction. The width of the band is ~1/5 of the uncut chip thickness (Fig. 17b), which is important for the estimation of the strain rate.

Thus, the model predicts the observed phenomena qualitatively and but the quantitative predictions need to be validated in future.

Figure 16. Simulated plastic strain during PMMA sectioning process, at the sectioning time of (a) 0.175 ms and (b) 0.195 ms.
Figure 17. Maximum strain distribution during sectioning of PMMA at an uncut chip thickness of 85 nm and sectioning speed of 10 mm/s, at a sectioning length of (a) 20 nm, and (b) 60 nm.

3.5 Nanosectioning of MWCNT/PMMA (Paper V)

PMMA reinforced by different contents of multi-walled carbon nanotubes (MWCNTs) were produced using a solution casting method. The dispersion of MWCNT in the matrix is shown in Fig. 18. Nanosectioning and Atkins’ model were used to assess the mechanical properties of the composites. The MWCNT contents and sectioning conditions are shown in Table 1. Nanoindentation was also used to validate nanosectioning results.

Table 1. Materials and nanosectioning conditions.

<table>
<thead>
<tr>
<th>MWCNT content (wt%)</th>
<th>Uncut chip thickness (nm)</th>
<th>Sectioning speed (mm/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.05</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.1</td>
<td>60, 80, 100,</td>
<td></td>
</tr>
<tr>
<td>0.2</td>
<td>120, 150, 200</td>
<td>1.0</td>
</tr>
<tr>
<td>0.5</td>
<td></td>
<td></td>
</tr>
<tr>
<td>1.0</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
A critical MWCNT content is perceivable in improving the shear yield stress of MWCNT/PMMA composites, as displayed in Fig. 19a. Below this content, the yield stress increased somewhat with the addition of MWCNTs, while beyond this content the yield stress showed a reducing trend, which is consistent with some previous studies [55–57]. The yield stress measured by nanosectioning was verified by the nanoindentation, which exhibited a similar MWCNT content dependence. The fracture energy of the composites was a few tens of J/m², and it showed an increasing trend as the MWCNT content increased, as shown in Fig. 19b. The estimated average interfacial fracture energy between MWCNT and PMMA is close to the values in [58,59].
3.6 Mechanical characteristics of CNF films (Paper VI)

Although the nanosectioning method was not employed in this part of the work, there is a common denominator in the underlying mechanisms in the polymer components of the composite, such as local yield, deformation and cracking at the sub-micrometre level.

Wood nanocellulose has been proposed for wound dressing applications because of its capability to form translucent films and aerogels with good liquid absorption capabilities (Fig. 20). In this study, the mechanical properties of three nanocellulose grades varying in the degree of nanofibrillation are assessed. The effect of nanofibrillation and of polyethylene glycol (PEG) addition (shown in Table 2), on the tensile strength, elongation and elastic modulus of the cellulose nanofiber (CNF) film is assessed in water and in phosphate-buffered saline (PBS). Fig. 21 shows the cross-sectional feature of the CNF_PEG film before and after the treatment in water.

![Figure 20](image) Nanocellulose-PEG film developed for wound management.

![Figure 21](image) Cross-sectional images of the CNF_PEG films in (a) 50% RH, and (b) in water.

<table>
<thead>
<tr>
<th>Film</th>
<th>Passes for homogenization</th>
<th>PEG (wt%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CNF01</td>
<td>1</td>
<td>-</td>
</tr>
<tr>
<td>CNF01_PEG</td>
<td>1</td>
<td>40</td>
</tr>
<tr>
<td>CNF02</td>
<td>2</td>
<td>-</td>
</tr>
<tr>
<td>CNF02_PEG</td>
<td>2</td>
<td>40</td>
</tr>
<tr>
<td>CNF03</td>
<td>3</td>
<td>-</td>
</tr>
<tr>
<td>CNF03_PEG</td>
<td>3</td>
<td>40</td>
</tr>
</tbody>
</table>

The tensile test results in Fig. 22 indicate that the fibrillation degree and addition of 40% PEG considerably affect the performance of the biocomposite dressings. In most cases, improving the fibrillation degree of the biocomposite yields an increase in the material strength and ductility. The PEG tends to decrease the strength and elastic modulus in wet conditions (water and PBS), while in most of the cases increases the strain to failure. This suggests that the PEG has a positive effect on the mechanical properties of the dressings, making them more flexible, ductile and potentially with higher
skin conformability. Such properties are considered most relevant for a wound management situation.

**Figure 22.** Tensile properties of the nanocellulose (PEG) bionanocomposite films in water and PBS conditions.

### 3.7 Summary

The estimated mechanical properties and the deformation behaviour of the studied materials presented in this thesis are summarised in Table 3.

**Table 3.** The mechanical properties and deformation behaviour of the studied materials.

<table>
<thead>
<tr>
<th>Material</th>
<th>Mechanical properties</th>
<th>Deformation</th>
<th>Presented in Paper</th>
</tr>
</thead>
<tbody>
<tr>
<td>PMMA</td>
<td>Fracture toughness: (~10\ J/m^2)</td>
<td>At and above the critical sectioning conditions, the deformation transmits from an homogeneous mode to a</td>
<td>I, II, III, IV</td>
</tr>
<tr>
<td>MWCNT/PMMA composites</td>
<td>Increases to 200 nm. The modulus decreases from 3 to 2.5 GPa as the speed varies from 0.25 to 10.0 mm/s.</td>
<td>Shear localised mode</td>
<td></td>
</tr>
<tr>
<td>------------------------</td>
<td>--------------------------------------------------------------------------------------------------</td>
<td>---------------------</td>
<td>---</td>
</tr>
</tbody>
</table>
| **Toughness:** 17 to 25 J/m² as MWCNT increases from 0 to 1.0 wt%.  
Shear yield stress: 97 to 103 MPa as MWCNT varying from 0-0.1 wt%, 90 to 60 MPa as CNT increases from 0.2-1.0 wt% | Ductile deformation | V |
| **Cellulose nanofiber (PEG) composites** | Tensile strength: < 0.2 MPa in water, 1-6 MPa in PBS.  
Strain to failure: 3-7 % in water, 4-14 % in PBS. | Elastic deformation | VI |
4 Conclusions

In this thesis, the deformation and fracture behaviour of a glassy polymer and nanocomposite during nanosectioning process are investigated by experiment and modelling. It starts with the measurement of the fracture toughness of glassy polymer under small scale deformation, then focuses on the effects of the sectioning condition (uncut chip thickness and sectioning speed) on the polymer deformation behaviour and on the surface mechanical properties of the glassy polymer, followed by the finite element modelling of the shear banding issue during sectioning, and ends with the exploration of the strengthening and toughening mechanisms in nanocomposites using sectioning and tensile test methods. The main investigation methods used in this work are presented in the following chart,

The main conclusions of this study are summarized as follows,

1. Measurement of the fracture toughness and yield stresses of materials

The instrumented ultramicrotome is able to measure the fracture toughness and shear yield stress of material at the nanoscale deformation. The fracture toughness of a commercial PMMA determined by nanosectioning was ~10 J/m², two orders of magnitude lower than the toughness determined at the macroscale [60], but was quite close to the theoretical surface energy
required to create new surfaces by breaking C-C bonds (1.5 J/m$^2$) [3]. The shear yield stress of the PMMA was approximately 110 MPa.

The fracture toughness of MWCNT/PMMA composites was measured by nanosectioning. It shows that the MWCNT can enhance the material toughness effectively, which increased from 17 to 25 J/m$^2$ as the MWCNT content increased from 0 to 1.0 wt%. The nanosectioning test reveals a critical MWCNT content in enhancing the strength of the composites. Below this content the composite strength showed an increase as the MWCNT content increased, while above this content the strength decreased.

2. Shear localisations

Periodic, wavy structures were formed during nanosectioning of PMMA, which significantly influenced the surface qualities and mechanical properties. Critical sectioning conditions (uncut chip thickness, sectioning speed) were revealed in the sectioning of PMMA, below which the created surfaces were flat and smooth, while above which periodic wavy structures formed on the surfaces. The average spacing between adjacent wavy structures was proportional to the uncut chip thickness. These wavy structures were identified as shear localisations originated from the shear deformation on the primary shear plane.

The shear localisations were associated with lower elastic moduli compared to the regions outside. The sectioning conditions had importance influence on the surface elasticity of PMMA. Increasing the sectioning speed led to a decrease in the surface elasticity. Increasing the uncut chip thickness decreased the surface elasticity but then improve the elasticity.

3. Formation mechanism of shear localisations

The analytical modelling of the stress variation in the primary shear zone and in the bulk material in front of the shear zone yields a critical speed for the onset of shear localisations in sectioning. This predicted speed is quite close to the experimental result of nanosectioning. The finite element analysis also confirms that localisations are formed during sectioning of PMMA. The material softening due to the adiabatic heating is the main reason for the formation of shear localizations. In fabrication of e.g. optical components, it would be useful to predict processing conditions to avoid formation of shear localisation defects.

4. Strengthening mechanism in nanocomposites

Incorporating MWCNT in PMMA with an appropriate content improved the material strength, while excessive addition decreased the material strength probably due to the MWCNT agglomeration.
Improving the fibrillation degree of the cellulose nanofibres led to an increase in the strength and ductility of the CNF biocomposite. The addition of PEG decreased the strength and elastic modulus in wet conditions, while in most of the cases increased the strain to failure.
5 Outlook

The work presented in this thesis reveals some interesting phenomena which take place during nanosectioning process, and these phenomena are analysed using the classic continuum mechanics. To fully understand the material deformation behaviour during nanosectioning process, more thorough studies need to be undertaken. There are several issues are worthy of investigations in future. From a scientific viewpoint, the following three challenges would deserve more efforts.

- Finite element analyse of sectioning of polymer matrix composites. The main challenge is the description of the filler-matrix interfaces during loading and debonding process. Improving the constitutive model of glassy polymers under dynamic load, especially at high rates, is also important. But this topic is highly related to the development of the experimental set-up since the tested strain rate is limited to the value of a few thousand now.

- Molecular dynamic simulation needs to carry out to understand the surface response during the sectioning at the nanoscale. The finite element simulation of the nanosectioning process using classic mechanics needs to be further validated and discussed.

- The measurement of the generated heat during sectioning process. To conduct temperature measurement at such a small scale is important to reveal the underlying mechanisms controlling the material behaviour, although this topic is highly dependent on the development of the miniature sensors.

From an application viewpoint, the methods and conclusions presented in this thesis also benefit the polymer manufacturing in industry. In future, the manufacturing of polymer components for optics and electronics using microsectioning and nanosectioning needs to consider the influences on the surface qualities caused by sectioning operation. The models presented in this thesis may be used to predict nanosectioning conditions to avoid damage formation for high quality products. In addition, more efforts need to be undertaken to improve the sectioning qualities, such as sectioning with lubrication fluid, cooling the work material, etc.
Sammanfattning på svenska


De viktigaste slutsatserna i avhandlingen, sammanfattas som följer:

1. Mätningar av brottsseghet och flytspänning hos material

Det är möjligt att uppskatta brottsenergi och skjuvflytspänning vid deformation på nanoskala med en ultramikrotom. Brottenergins hos en kommersiell PMMA uppmättes på detta sätt till ~10 J/m², vilket är nära den teoretiska ytenergin för att bryta de kovalenta C-C-bindningarna (1.5 J/m²), men två storleksordningar mindre än vad som uppmäts makroskopiskt vid stan-
dardiserad brottmekanisk provning. Skjuvflytspännningen hos PMMA upp-
skattades till c:a 110 MPa.

Även brottsenergin hos kompositer bestående av flervågiga kolnanorör i
PMMA-matris uppmättes. Resultatet var att kolnanorören ökade brottenergin
på materialet från 17 till 25 J/m², då andelen förstärkning ökades från 0 till
1.0 viktsprocent. Vidare upptäcktes ett tröskelvärde för koncentration för-
stärkning för att höja styrkan hos kompositmaterialet. Under denna kritiska
mängd ökade styrkan med ökad mängd kolnanorör, men minskade istället
vid högre koncentrationer.

2. Lokaliserad skjuvning

En periodisk och vågliknande yttextur skapades vid snittning av PMMA,
vilket visat sig påverka dess mekaniska ytegenskaper. Kritiska förhållanden
(snittjockey och snitthastighet) kunde kostateras, vilket avgör om ytan får
periodiska mönster eller blir jämn och slät. Den genomsnittliga perioden av
den vågiga formen visade sig vara proportionell med snittjockeyden. Dessa
periodiska former identifierades som lokalisera skjuvflytning.

Områden som uppvisar lokaliserad skjuvdeformation uppvisade en lägre
elasticitetsmodul jämfört med övriga intakta områden. Förhållandena vid
snittning (hastighet, tjocklek, temperatur osv.) har en stor inverkan på snitty-
tan i PMMA. En ökad snitthastighet ledde till en minskad elasticitetsmodul
på ytan. En ökning av snittjockeyden visade sig först leda till en minskning
av ytans styvhet, och därefter till en ökning.

3. Mekanismer bakom lokaliserad skjuvflytning

En analytisk modell har använts för att bestämma spänningstillståndet i
det primära skjuvområdet och i materialet framför skjuvområdet. En kritisk
snitthastighet för aktivering av skjuvflytning kunde påvisas. Den beräknade
kritiska hastigheten är nära den hastighet som bestämdes experimentellt.
Numeriska simuleringar med finita element bekräftar att lokalisering av
skjuvning sker vid snittning av PMMA. Den största anledningen till lokali-
serad skjuvning är att lokala temperaturhöjningar vid plastisk deformation,
da materialet mjuknar påtagligt.

4. Förstärkningsmekanismer i nanokompositer

Förstärkning av PMMA med kolnanorör i lämplig mängd visade sig leda
till en skönjbara högre hållfasthet i nanoskala, medan en för stor mängd
ledde till en minskad styrka, vilket sannolikt beror på aggregering av kol-
nanorören.
En ökad fibrilleringsgrad av cellulosananofibriller visade sig leda till en högre dragstyrka och duktilitet i en biokomposit tänkbar som sårskydd. Polyetenglykol verkar sänka styrkan och styvheten i vått tillstånd, medan töjbarheten ökades.
References


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