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Technical Note

The effect of clamping conditions on tearing energy estimation for highly stretchable materials

L. Bernardi [a], E. Mazza [a,b], A.E. Ehret [a,b,*]

[a] ETH Zurich, Institute for Mechanical Systems, 8092 Zürich, Switzerland
[b] Empa, Swiss Federal Laboratories for Materials Science and Technology, 8600 Dübendorf, Switzerland

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ABSTRACT

A typical measure for the toughness of highly stretchable materials is the energy for tearing, determined in mode I fracture tests on wide specimens with a lateral cut. In the present paper we show that, when slippage occurs in the grips that hold the specimens, the classical analysis of this test leads to an overestimation of the tearing energy, if sample stretches are determined from the grip displacements. Unlike for elastic properties, the use of local strain data retrieved by an optical measurement system does not remedy this issue but, *vice versa*, underpredicts the tearing energy. Here we propose a simple *post hoc* correction of the measured tearing energy, and an improved clamp design to prevent slippage. Applied to a commercial acrylic elastomer, the corrected tearing energy was independent of the occurrence of slippage and consistent with the result obtained with the new clamping system as well as values reported in literature.

1. Introduction

For many applications of soft, highly deformable materials, their fracture properties, particularly the ability to resist existing defects without catastrophic failure, is a critical factor. Examples of the multitude of industrial applications of natural and synthetic elastomers are seals, belts, tubes and dampers [1] or dielectrics in soft generators and actuators [2,3]. Silicone elastomers have a wide range of uses in medical devices and implants including artificial urinary sphincter [4], external shell of breast implants [5], contact lenses [6] and skin supports [7]. Leslie et al. [8] underline that little is known about fracture properties of silicones used in medical implants despite this is an important aspect in the design of the device itself. One example concerns finger joint implants [8,9]. Bass et al. [10] report that 45% of implants fracture after 3 years from implantation, possible causes for fracture initiators are small cracks created by spurs from the finger bones [11,12]. Recently, great effort has been put into increasing stiffness and fracture toughness of elastomers, for example introducing prestretched chains that break and dissipate energy before the failure of the material [13] and particularly tough hydrogels have been developed [14–16] with applications in contact lenses, scaffolds for tissue engineering, wound dressing, drug delivery or artificial tissues [17].

The resistance to break from an existing defect is characterised by the rather generic term [18] "fracture toughness" embracing different metrics that quantify some experimental fracture properties. The typical experimental measure of fracture toughness for materials that experience large elastic deformations is the "characteristic energy for tearing" Γ defined in

* Corresponding author at: Empa, Swiss Federal Laboratories for Materials Science and Technology, Überlandstrasse 129, 8600 Dübendorf, Switzerland.
E-mail address: ehret@imes.mavt.ethz.ch (A.E. Ehret).

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Nomenclature

B	initial sample width
C	initial crack length
dC	crack increment
dl	increase of grip distance
F	measured force
L	initial sample length
L^*	updated reference length
L_c^*	updated reference length at crack propagation
l	current sample length
l_c	length at which crack propagates
P	nominal stress
$\mathscr{R}_{PS}$	region of the sample in a state of pure shear
T	initial sample thickness
U	elastically stored energy
W	elastically stored energy density
Γ	characteristic energy for tearing
Γ^*	slippage-corrected tearing energy
$\bar{\Gamma}$	tearing energy when local stretch is used
γ_c	correction factor
λ	nominal longitudinal stretch
λ_c	nominal longitudinal stretch at crack propagation
$\bar{\lambda}$	'true' local stretch
$\bar{\lambda}_c$	'true' local stretch at crack propagation
ζ	local stretch in intact sample

the seminal work by Rivlin and Thomas [19]. It is typically determined from either the "trousers test" [20,21] or using test-pieces with high width-to-length ratio and a lateral cut, which is likewise larger than the sample length L. For incompressible materials, the latter test is known as "pure shear" tear test [19], and in order to determine the tearing energy, the specimen is extended in a displacement-controlled manner until a crack propagates from the cut tip. This occurs at a certain current length $l = l_c$, measured as the distance between the grips holding the sample. To calculate Γ from these experiments, according to the theory [19], it is assumed that within a certain zone of the specimen that is in a state of pure shear, the change of sample length from L to l is an elastic process, i.e. all work spent to elongate the sample leads to elastically stored energy. Upon crack growth, a small volume element of this zone is unloaded and the stored energy is released to propagate the crack. It is well known that if the material is visco- or inelastic, the tearing energy determined in this way overestimates the actual fracture energy since a part of the work expended on the system has been dissipated and hence cannot contribute to driving the crack [18,22,23]. Nevertheless, for viscoelastic materials, the experiments can be performed at low strain rate to reach a quasi-static regime, or an 'apparent' tearing energy can be defined for a fixed strain rate [24]. Misestimation of the tearing energy, however, is not only caused due to energy that is dissipated by the material internally, but any other component of the testing system that is between the force transducers. This concerns in particular the work expended due to changes occurring in clamping conditions (e.g. slippage). The calculation of the stored energy assumes perfect clamping so that the distance between the clamps is equal to the current free length of the specimen throughout the test. This is generally an idealisation, and for mechanical materials characterization, the use of optical measurement techniques, e.g. digital image correlation, to determine the in-plane strain field has become a standard in most laboratories. As will be shown in this paper, the use of these 'local' data alone is not sufficient to determine the correct fracture energy. However, in combination with the nominal data a simple correction factor can be derived that allows to determine a slippage independent fracture energy.

In the present work, we present this correction in application to VHB^{TM}4905 (3M^{TM}, Saint Paul, Minnesota, USA), a highly deformable acrylic elastomer. The fracture characteristics of this material were recently investigated in detail [24] by use of an alternative clamping technique [25] that prevents slippage, and the value of 1.5 kJ/m^2 obtained by Pharr et al. [24] serves as a reference value for the present study.

2. Materials and methods

2.1. The effect of slippage

In the "pure shear" tear test [19] a rectangular specimen with large width-to-length ratio and a lateral cut is elongated in a displacement-controlled manner until, at a certain elongation corresponding to the grip distance l_c, a crack propagates from the tip of the cut (cf. Fig. 1). A region $\mathscr{R}_{PS}$ of the sample in a reasonable distance from both the tip of the cut and

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