The final goal of this PhD-work is an efficient and user-friendly finite element modelling strategy targeting an industrial available package opening application. In order to reach this goal, different experimental mechanical and fracture mechanical tests were continuously refined to characterize the studied materials. Furthermore, the governing deformation mechanisms and mechanical properties involved in the opening sequence were quantified with full field experimental techniques to extract the intrinsic material response. An identification process to calibrate the material model parameters with inverse modelling analysis is proposed. Constitutive models, based on the experimental results for the two continuum materials, aluminium and polymer materials, and how to address the progressive damage modelling have been concerned in this work. The results and methods considered are general and can be applied in other industries where polymer and metal material are present.

This work has shown that it is possible to select constitutive material models in conjunction with continuum material damage models, adequately predicting the mechanical behaviour in thin laminated packaging materials. Finally, with a slight modification of already available techniques and functionalities in a commercial general-purpose finite element software, it was possible to build a simulation model replicating the physical behaviour of an opening device. A comparison of the results between the experimental opening and the virtual opening model showed a good correlation.

The advantage with the developed modelling approach is that it is possible to modify the material composition of the laminate. Individual material layers can be altered, and the mechanical properties, thickness or geometrical shape can be changed. Furthermore, the model is flexible and a new opening design with a different geometry and load case can easily be implemented and changed in the simulation model. Therefore, this type of simulation model is prepared to simulate sustainable materials in packages and will be a useful tool for decision support early in the concept selection in technology and development projects.
Mechanics and Failure in Thin Material Layers
Towards Realistic Package Opening Simulations

Eskil Andreasson
Mechanics and Failure in Thin Material Layers
Towards Realistic Package Opening Simulations

Eskil Andreasson

Doctoral Dissertation in Mechanical Engineering

Department of Mechanical Engineering
Blekinge Institute of Technology
SWEDEN
Mechanics and Failure in Thin Material Layers

Towards Realistic Package Opening Simulations
This thesis is dedicated to my family

Monika, Vera, Lukas and Simon, I love you!
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This PhD-thesis is the result of my doctoral studies conducted at the department of Mechanical Engineering at Blekinge Institute of Technology (BTH), in Karlskrona, Sweden. The work has been included in the KKS-profile Model Driven Development and Decision Support – “MD3S” and has been financed by Tetra Pak®. All the involved persons devoted and dedicated colleagues and managers and especially my sponsors at Tetra Pak® are all greatly acknowledged.

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All the valuable discussions and great work that has been achieved within all Master thesis projects that I have been involved in are highly appreciated. The benefits with young people are that they are often totally unafraid of jumping into an “ocean” of unknown issues and obstacles that arise in industry. Different personalities, nationalities, genders, backgrounds have taught me a lot. All this work has really pushed and motivated me to search for more knowledge and techniques to be able to answer all the questions that have been raised during this challenging, tough and rewarding applied research work.

Knowledgeable colleagues, professors, material experts and personnel in workshops at different laboratories and large-scale facilities are all acknowledged. The daily discussions with my great colleagues/friends have been instrumental for completion of this work. I am thankful for the working climate and positive energy which has been essential for succeeding to increase the use of an engineering sound, realistic and simple to use simulation approach without the need of bursting.

Finally, I would like to thank all my family and friends for all the unconditional love, patience and support during this tough, fun and tremendous learning and time-consuming project and journey of my life.

Eskil Andreasson, 2019-01-01, Font de Sa Cala - Mallorca
A special thanks to all of you that have financed my PhD-thesis, supported me, inspired me and helped me in different ways to complete this work!

Abstract

The final goal of this PhD-work is an efficient and user-friendly finite element modelling strategy targeting an industrial available package opening application. In order to reach this goal, different experimental mechanical and fracture mechanical tests were continuously refined to characterize the studied materials. Furthermore, the governing deformation mechanisms and mechanical properties involved in the opening sequence were quantified with full field experimental techniques to extract the intrinsic material response. An identification process to calibrate the material model parameters with inverse modelling analysis is proposed. Constitutive models, based on the experimental results for the two continuum materials, aluminium and polymer materials, and how to address the progressive damage modelling have been concerned in this work. The results and methods considered are general and can be applied in other industries where polymer and metal material are present.

This work has shown that it is possible to select constitutive material models in conjunction with continuum material damage models, adequately predicting the mechanical behaviour of thin laminated packaging materials. Finally, with a slight modification of already available techniques and functionalities in a commercial general-purpose finite element software, it was possible to build a simulation model replicating the physical behaviour of an opening device. A comparison of the results between the experimental opening and the virtual opening model showed a good correlation.

The advantage with the developed modelling approach is that it is possible to modify the material composition of the laminate. Individual material layers can be altered, and the mechanical properties, thickness or geometrical shape can be changed. Furthermore, the model is flexible and a new opening design with a different geometry and load case can easily be implemented and changed in the simulation model. Therefore, this type of simulation model is prepared to simulate sustainable materials in packages and will be a useful tool for decision support early in the concept selection in technology and development projects.
Keywords: aluminium foil, FEM, LDPE, localisation, necking, polymer, progressive damage, semi-crystalline, simulation
Sammanfattning


Simuleringsmodellerna som kontinuerligt förbättras och utvecklas i detta arbete har som slutmål att prediktera öppningsförloppet samt öppningsbarhet av dryckesförpackningar. Modellerna ska vara tillförlitliga, användarvänliga och möjliga att använda som beslutsstöd inom industriella applikationer. Tre beståndsdelar är viktiga i denna typ av simuleringsmodeller; geometri, materialmekanik och laster/randvillkor. Respektive del måste genomarbetas för att kunna uppnå en tillförlitlig och realistisk modell. Mekanisk och brottmekanisk materialkarakterisering för tunna materialskikt har varit en viktig del i detta arbete. De existerande experimentella teknikerna och metoderna som använts har kontinuerligt uppdaterats och förbättrats. Vid behov har nya tillgängliga experimentella tekniker och testmetoder införts. Detta experimentella arbete är en viktig informationskälla till modelleringarbetet i den virtuella miljön. En effektiv kalibreringsmetodik har utarbetats i detta arbete för att identifiera materialparametrar som sedan används i de virtuella materialmodellerna för att beskriva respektive materialskikts mekanik.

Denna avhandling har slutligen påvisat att det med stor tillförlitlighet går att prediktera öppningssekvensen med hjälp av en virtuell öppningssimulering av en dryckesförpackning. De experimentella metoderna och kalibreringsteknikerna som utvecklats under arbetet kan med fördel överföras och användas i andra industrier och tillämpningar där metall eller polymer förekommer.
List of included papers

This PhD-thesis is organized as a compilation thesis, meaning that it consists of five scientific papers. Furthermore, the PhD-thesis includes an introduction and overview of the whole PhD-project. The work is based to a large extent on the following five scientific research papers prepared by the author and co-workers, which are published, presented or in the process of being published:

**Paper I – Thin packaging material layer, Aluminium foil**

*Simulation of Thin Aluminium-foils in the Packaging Industry*

Eskil Andreasson, Tommy Lindström, Britta Käck, Christoffer Malmberg, Ann-Magret Asp  
AIP Conference Proceedings 1896, 160014 (2017),  
20th International ESAFORM Conference on Material Forming, Dublin, Ireland  
Published by the American Institute of Physics  
https://doi.org/10.1063/1.5008189

**Paper II – Manufacturing process and properties of Al-foil**

*On the stiffness tensor in AA8079 at small and intermediate strains*

Eskil Andreasson, Wureguli Reheman, Per Ståhle, Sharon Kao-Walter (2019)  
submitted for publication

**Paper III – Micro mechanics in two thin packaging materials**

*Micro-mechanisms of a laminated packaging material during fracture*

Eskil Andreasson, Sharon Kao-Walter, Per Ståhle (2014)  
Published in Engineering Fracture Mechanics 127, 2014, pages 313–326  
http://dx.doi.org/10.1016/j.engfracmech.2014.04.017

**Paper IV – Injection moulded polymer material, continuum**

*Anisotropic Elastic-Viscoplastic Properties at Finite Strains of Injection-Moulded Low-Density Polyethylene*

Published in Experimental Mechanics (2018) 58: 75, Springer  
https://link.springer.com/article/10.1007/s11340-017-0322-y

**Paper V – Putting it all together, opening device application**

*Advancements in package opening simulations*

Eskil Andreasson, Joel Jönsson (2014)  
Published in Procedia Materials Science 3, 2014, pages 1441–1446  
Published in the proceedings of the 20th European Conference on Fracture (ECF20)  
Own contribution in the appended papers

The author of this thesis has taken the main responsibilities for the planning, preparation and writing of Paper I, III and V. Paper II was planned and written together with the co-authors. Paper IV was planned and written together with Martin Kroon and the co-authors. Furthermore, experimental tests combined with the development of theories and numerical simulations have in all papers been done in close collaboration with the co-authors.
Related Work

The following publications are related to the work presented in this thesis but have not been included or appended in this printed PhD-thesis:

*Trouser tear testing of thin anisotropic polymer films and laminates*
Md Shafiqul Islam, Eskil Andreasson and Sharon Kao-Walter (2019)
Submitted for publication

*Ab initio and classical atomistic modelling of structure and defects in crystalline orthorhombic polyethylene: twin boundaries, slip interfaces, and nature of barriers*
Pär A. T. Olsson, Elisabeth Schröder, Per Hyldgaard, Martin Kroon, Eskil Andreasson and Erik Bergvall (2018)
Polymer, Elsevier, Volume 121, Pages 234-246

*All-atomic and coarse-grained molecular dynamics investigation of deformation in semi-crystalline lamellar polyethylene*
Polymer, Elsevier, Volume 153, Pages 305-316

*Experimental and numerical assessment of work of fracture in injection-moulded low-density polyethylene*
Engineering Fracture Mechanics, Volume 192, Pages 1-11

*Numerical Analysis of Anisotropic stiffness of thin Al-foil in multiple material directions based on Experiments*
Wureguli Reheman, Per Ståhle, Eskil Andreasson and Sharon Kao-Walter (2017)
Conference: 30th Nordic Seminar on Computational Mechanics (NSCM30), Technical University of Denmark (DTU), Copenhagen

*Powerful Modelling Techniques in ABAQUS to Simulate Failure of Laminated composites*
DOI: 10.13140/RG.2.2.28600.03845
Research report ISSN 1103-1581; 2016:01, BTH, Karlskrona, Sweden
Realistic Package Opening Simulations - An Experimental Mechanics and Physics Based Approach
Eskil Andreasson (2015)
Licentiate Dissertation, Series No. 2015:02, Department of Mechanical Engineering, Blekinge Institute of Technology, BTH, Sweden

SEM observations of a metal foil laminated with a polymer film
Nasir Mehmood, Eskil Andreasson and Sharon Kao-Walter (2014)
DOI: 10.1016/j.mspro.2014.06.232
Conference: ECF20: Procedia materials science, Trondheim, Norway

Trouser tear tests of two thin polymer films
Eskil Andreasson, Nasir Mehmood and Sharon Kao-Walter (2013a)
13th International Conference on Fracture (ICF), June 16–21, Beijing, China

Integrating Moldflow and Abaqus in the Package Simulation Workflow
Eskil Andreasson, Leo Persson, Henrik Jacobsson and Johan Nordgren (2013b)
Conference: SCC2013, 2013 SIMULIA Community Conference in Vienna, Austria

Deformation and Damage Mechanisms in Thin Ductile Polymer Films
Eskil Andreasson, Joel Jönsson, Martin Sandgren and Paul Håkansson (2013c)
NAFEMS NORDIC Seminar: Improving Simulation Prediction by Using Advanced Material Models, November 5-6, Lund, Sweden

Is it possible to open beverage packages virtually?
-Physical tests in combination with virtual tests in Abaqus
Proceedings of the SIMULIA Community Conference, Providence, USA
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### Abbreviations

<table>
<thead>
<tr>
<th>Abbreviation</th>
<th>Description</th>
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<tbody>
<tr>
<td>Al</td>
<td>Aluminium</td>
</tr>
<tr>
<td>ALE</td>
<td>Arbitrary Lagrangian Eulerian</td>
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<td>BC</td>
<td>Boundary Conditions</td>
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<td>BIB</td>
<td>Broad Ion Beam</td>
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<td>BTH</td>
<td>Blekinge Tekniska Högskola</td>
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<tr>
<td>CAD</td>
<td>Computer Aided Design</td>
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<tr>
<td>CEL</td>
<td>Coupled Eulerian Lagrangian</td>
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<td>CD</td>
<td>Cross Direction</td>
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<td>CDM</td>
<td>Continuum Damage Mechanics</td>
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<tr>
<td>CP</td>
<td>Crystal Plasticity</td>
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<td>CT</td>
<td>Computerized Tomography</td>
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<td>DCT</td>
<td>Diffraction Contrast Tomography</td>
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<td>DD</td>
<td>Diagonal Direction</td>
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<td>DFT</td>
<td>Density Functional Theory</td>
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<td>DIC</td>
<td>Digital Image Correlation</td>
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<td>DSP</td>
<td>Digital Speckle Pattern</td>
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<tr>
<td>DVC</td>
<td>Digital Volume Correlation</td>
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<tr>
<td>EBSD</td>
<td>Electron Back Scattering Diffraction</td>
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<tr>
<td>EC</td>
<td>Extrusion Coating</td>
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<tr>
<td>ECF</td>
<td>European Conference of Fracture</td>
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<tr>
<td>ESS</td>
<td>European Spallation Source</td>
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<tr>
<td>FE</td>
<td>Finite Element</td>
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<tr>
<td>FEM</td>
<td>Finite Element Method</td>
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<td>FIB</td>
<td>Focal Ion Beam</td>
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<tr>
<td>GUI</td>
<td>Graphical User Interface</td>
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<tr>
<td>ICF</td>
<td>International Conference of Fracture</td>
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<tr>
<td>IM</td>
<td>Injection Moulding</td>
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<tr>
<td>KKS</td>
<td>Kunskap och Kompetens Stiftelsen</td>
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<tr>
<td>LDPE</td>
<td>Low Density Polyethylene</td>
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<tr>
<td>LOM</td>
<td>Light Optical Microscopy</td>
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<td>MD</td>
<td>Machine Direction</td>
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<td>MD</td>
<td>Molecular Dynamics</td>
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<td>MD3S</td>
<td>Model Driven Development and Decision Support</td>
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<tr>
<td>SAXS</td>
<td>Small Angle X-ray Scattering</td>
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<td>SEM</td>
<td>Scanning Electron Microscopy</td>
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<td>PhD</td>
<td>Philosophiæ Doctor</td>
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<tr>
<td>PLH</td>
<td>Pre-Laminated Hole</td>
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<tr>
<td>PM</td>
<td>Packaging Material</td>
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<td>RD</td>
<td>Rolling Direction</td>
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<td>TD</td>
<td>Transverse Direction</td>
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<tr>
<td>VPL</td>
<td>Virtual Package Laboratory</td>
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<tr>
<td>WAXS</td>
<td>Wide Angle X-ray Scattering</td>
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<tr>
<td>XCT</td>
<td>X-ray Computerized Tomography</td>
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<tr>
<td>XFEM</td>
<td>eXtended Finite Element Method</td>
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<tr>
<td>XRD</td>
<td>X-ray Diffraction</td>
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1. Introduction

During the last decades the Finite Element Method (FEM) has been more frequently introduced and used in the packaging industry as a complement to Computer Aided Design (CAD). Computer-based simulation models with realistic results, often referred as “virtual twins”, are used early in the development process today to ease and facilitate development and decision making in the concept and design phase, cf. Cadge (2015). Furthermore, the simulation models are predictive at a subsystem level, macroscopic application level and lately on a microscopic level. Hence multiple length scales are nowadays involved and included in the simulation models.

1.1 Background and motivation

Due to a shift in peoples’ life style on-the-go consumption with new and additional functionality of opening devices have increased in the packaging sector, cf. Skoda (2017). Most of the packaging materials consists of several material layers; i.e. polymers, aluminium-foil and paperboard laminated together as shown in Figure 1, cf. Bolzon (2015). New composites and types of materials in combination with more complex functionalities included in the package openings calls for a simulation-based decision support in product development. All these changes motivate attention to more sophisticated simulation models and material descriptions. Therefore, more experimental data are needed as input and verification of the models. The fracture processes that occur during the opening sequence of a package is intended. Moreover, the damage initiation and propagation leading to complete failure should be controlled. Therefore, an increased knowledge of how the different packaging material layers react to the prevailing loading scenario in a real case situation is necessary, cf. Andreasson (2012). Increased functionality in the design of the opening devices calls for a systematic and better understanding of the opening sequence and the process in general. Simultaneously the experimental techniques, hardware and software capabilities are constantly increasing and improved. More and smaller geometrical details are embedded and accounted for in the simulation models. These enhanced functionalities push for an increased knowledge of the microstructure and evolution of the structure in the studied materials. A virtual packaging material description have emerged continuously during this work. Moreover, scenario-based modelling and sensitivity analysis are performed in a more efficient way.
Increased and better understanding of the mechanical performance, noise parameters and environmental effects is continuously growing. Therefore, how the individual packaging material layers are affected by these factors is a prerequisite to be able to do realistic predictions of the package performance with the simulation models.

The packaging material layers are often described in an isotropic, homogenized non-linear elasto-plastic framework in the simulation models today. This simplification still facilitates accurate results in macroscopic applications. However, to enhance the current models and to enter next level of accuracy of realistic models more refined details must be accounted for and included in the simulation models.

An injection moulded opening device, further presented in Andreasson et al. (2012) and in Paper V, was chosen as a reference opening device in this PhD-thesis. During the different stages of this specific opening process, when cutting through a thin laminated membrane, several of the phenomena and mechanisms from a mechanical point of view are triggered. Four mechanisms are involved, important to understand and to accurately quantify in experiments for model parameter identifications:

1. **Mechanical material behaviour** - deformation of the membrane
2. **Damage initiation material behaviour** - cutting of the membrane
3. **Bond strength** - i.e. material interfaces with traction law between the material layers
4. **Contact/interaction** - friction between the different parts

The focus herein and in the included papers has been to further investigate the first two topics highlighted above. In Andreasson et al. (2012) it was evident that a dedicated research work related to an increased understanding of the fundamental
mechanics in the individual packaging material layers and interaction between the
different material layers was required. Theoretical aspects, experimental techniques,
experimental tests, material mechanics and especially a transfer of the experimental
observations and findings into a simulation model, e.g. material models, at different
length scales ranging from micro-mechanical models to more macroscopic and
application models have been the corner stones of this PhD-thesis.

1.2 Objective and vision

An experimental campaign of mechanical characterization in thin material layers and
a finite element modelling strategy targeting package opening simulations has been
developed during this work. This information and virtual models will be used to
predict the opening performance early in the concept selection and act as decision
support in projects in the future. The focus has been on a combined physical/virtual
test procedure for mechanical characterization and material model calibration of thin
packaging materials. Freestanding and laminated material layers of packaging
materials has been studied experimentally in different material directions and loading
conditions at different temperatures and strain rates. The governing mechanical
material properties involved in the opening performance need to be identified and
experimentally quantified. An accurate and reliable continuum and damage material
model needs to be utilized in the finite element model to be able to predict the
intended progressive damage behaviour occurring during the opening sequence. The
following statements describe the vision of the project:

*Development of sustainable and easy-open opening devices can be further enhanced
by an increased knowledge of the involved and activated deformation mechanisms and
the individual mechanical properties of each packaging material layer in the
packaging material laminate.*

* A fundamental understanding of the materials and package manufacturing process,
process induced properties, microstructure, deformation and the accompanying
fracture process is considered when improving existing and developing new opening
and closure devices.*

*Reliable virtual engineering tools, i.e. simulation models – “virtual twins” in
conjunction with qualitative physical testing, will guide the packaging materials,
opening and closure development for the future in the digital era.*
1.3 Outline of the thesis

The work is composed by three components; (i) Manufacturing Processes, (ii) Material Structure and Material Mechanics and finally the (iii) Simulation Strategy that is input and used in the opening simulation model. All these three components are important in a physically rooted, engineering sound and simple to use simulation model with realistic results. The next sections present a general overview of the project, linking all the work conducted and potential enhancements needed. The reader is provided with further guidance to find a more thorough explanation in the included papers or related works and references. Theoretical aspects are discussed in Papers I-IV. Paper V is summarizing and utilizing all the built knowledge.

This PhD-thesis is organized as follows: Section 2 describes the manufacturing processes involved during the creation of material layers. Sections 3 - 4 introduces the two different materials studied and the inherited mechanical properties induced by these manufacturing processes in the layers at different length scales. Sections 5 - 6 describes the experimental tests performed and how the virtual models have been developed and utilized. Finally, Sections 7 - 8 summarize the five papers and presents the conclusions of the work and Section 9 is an outlook.

1.3 Limitations

Paperboard is an important part in the packaging material composition and has not been included in this specific work. However, there has been a tremendous effort and work on paperboard that has inspired significantly to this PhD-work. The interested reader is referred to Tryding (1996), Dunn (2000), Xia et al. (2002), Mäkelä et al. (2003), Nygård et al. (2009), Beex et al. (2009), Giampieri (2011), Lindström (2013), Borgqvist et al. (2014), Linvill (2014) to name a few. Moreover, adhesion or the interfaces between the different material layers has not been strictly focused in this work. It is left outside the scope of this PhD-thesis. Different length scales have been studied in this work, primarily starting from a few μm and above. The smaller length scales probed with various techniques such as scattering techniques with for instance X-rays referred as SAXS/WAXS or XRD has not been included in this work, the interested reader is referred to Schmacke (2010) and Björn (2018). Furthermore, simulation models complementing the available measurements at this atomistic length scale are interesting and there is an ongoing work within this field referred as Molecular Dynamics and DFT, cf. Yeh (2017) and Olsson et al. (2018).
2. Manufacturing Processes

A laminated packaging material often consists of three different materials; paperboard, polymer and aluminium-foil. In addition to these material layers polymer parts e.g. openings, caps and tops are often attached to the package. In this PhD-thesis the focus is on the polymer and aluminium-foil materials. The paperboard material is not included in this scope. Therefore, the manufacturing processes associated with the production of polymer and aluminium materials are of interest. For instance, in Toft et al. (2002) extrusion coated films and the influence of processing conditions were studied. Konijnendijk et al. (2007) have studied the injection moulding process. Moreover, information on the process induced structure and the inherited mechanical performance created by such manufacturing processes is found in Nordgren et al. (2012), Andreasson et al. (2014). Recently, Björn (2018) has studied the process induced semi-crystalline polymer structure inherited from the injection moulding process on the polymer crystal structure and Olsson et al. (2018) MD-modelled these semi-crystalline polymer systems at an atomic length scale.

Manufacturing processes are the steps where the raw material is refined. The final material is created during this sequence. Therefore, geometry, surfaces, internal structures, microstructure and mechanical performance is governed by the type of process that is selected and the associated process settings, which will be described more in Papers I-II for aluminium foil and in Paper IV for injection moulded polymer material. Materials in many industries get more advanced, for instance combination of different materials acting as a composite. Prediction of the mechanical behaviour with computer simulations calls attention to the coupling between the raw material, manufacturing process and link these attributes to the performance.

The three different material manufacturing processes utilized in this work are represented with schematic drawings in Figure 2a), b) and c). Double twin rolling of aluminium foil, extrusion coating and injection moulding of polymer material presented in the picture represents the processes of interest. It was found that the same “raw”-material can create different final products depending on process conditions and individual process settings. Furthermore, within the same class of materials mechanical properties of polymers can show from brittle like failure to very ductile mechanical behaviour when exposed to mechanical load. These large differences can be attributed to the manufacturing process technique, or process conditions or subsequent process history that the material is exerted to after
production. Therefore, a good understanding of the process and the corresponding process settings utilized during the manufacturing is fundamental to gain. This can be achieved by studying the material concerned at various length scales with complementary and different experimental techniques revealing this local information. Information of how the material is composed and if different material phases, elements or structures are present in the material is possible to extract with experimental techniques. Furthermore, how the material reacts to mechanical load, what mechanisms and geometrical changes that are involved and how the structure evolves during these conditions can be studied. Simulation models can be applicable to the material production as well, for instance aluminium foil production can be simulated including both the structure and the surfaces in the simulation model. Engler et al. (2014) has studied the development of the surface topography created in the manufacturing process to be able to control and, eventually, improve the properties of aluminium foil. The aluminium foil is concerned in Papers I, II and III.

Figure 2. a) Double twin-rolling of aluminium foil b) Extrusion coating of polymer materials c) Injection moulding of polymer materials. Courtesy of Manufacturing Guide. 
3. Material Mechanics and Damage Mechanics

A laminated packaging material consists of three different materials; paperboard, polymer and aluminium-foil. At first, the mechanics of each individual material layer was studied to predict the mechanical behaviour of the complete composite. Moreover, the microstructure inherited by respective manufacturing processes and the microstructure evolution that is ongoing during the deformation process governs the macroscopic mechanical response in the constituents. The microstructure may be altered during the different manufacturing steps with speed, mechanical load and the individual process settings. Despite the different origin of the three different materials present in the packaging material, a lot of similarities were discovered. Although, large differences can be found in the length scales involved in the microstructures. The different material groups are represented with Scanning Electron Microscopy (SEM) micrographs in Figure 3, to the left and in the middle polymer films are illustrated. In the picture to the right, a SEM-micrograph of a cross section in thin aluminium foil, is illustrated.

Borgqvist et al. (2014), Nygårds et al. (2009), Beex (2009), Mäkelä et al. (2003), Xia et al. (2002), and Tryding (1996) have studied the mechanics of paperboard materials. These prior works related to paperboard have inspired and the knowledge have been transferred to the aluminium foil and the polymer materials studied in this work. Several recent research studies have focused on the mechanical behaviour and deformation process in semi-crystalline polymer materials and aluminium foil used in the packaging industry; e.g. Shahmardani (2018), Kao-Walter (2002, 2004, 2011), Bolzon et al. (2012, 2015, 2017, 2018), Mehmood et al. (2012), Jönsson et al. (2013), Xiang (2005) and Nordlund et al. (2014).

![Figure 3. SEM-pictures of the two different packaging materials; extrusion coated polymer films and aluminium foil. Micrographs by Nasir Mehmood](image-url)
Typical mechanical response graphs of standard experimental tests are presented in Figure 4 for the three different material groups described. However, several research questions remain in the area of thin semi-crystalline polymer films and thin aluminium foils. Therefore, the early stage of the work was focused on these two material groups. One challenge with aluminium foil is the thin thickness, approximately 10 µm in the studied applications when used in the packaging industry. This dimension is in the length scale of one or a few grains through the thickness. Aluminium foil is concerned and summarized in Papers I, II and III and has been further investigated by Shahmardani (2018), Bolzon et al. (2012, 2015, 2017, 2018), Käck et al. (2015), Larsson (2017) and Duse et al. (2017).

To be able to accurately predict the damage evolution in the laminated material, the mechanical response in different loading scenarios of the aluminium foil and the polymer films are at first studied as separate material layers in Papers I, III and IV. To form a well-defined basis for the investigation, centre-cracked panels exposed to in-plane uniaxial tensile mode I loading are further analysed in Paper III. The interfaces between the material layers, i.e. the bond strength, have not yet been thoroughly addressed in this study. However, peel test studies have been made together with Bruce et al. (2013) and Postlind et al. (2016).

Polymer materials can resist high deformations/strains before failure as described in Papers III and IV. Strain hardening is present and is one of the main effects involved in the latter part of the deformation process of the studied polymer film during the opening process. Thus, the complete experimentally measured mechanical material response curve needs to be accurately transferred into the numerical material model.
Morphological evolution and geometrical changes, such as necking in the width and thickness direction, are included phenomena in the deformation process and has to be accounted for. Moreover, these microscopic events need to be accounted for in the macroscopic material model description in the virtual models. However, the macroscopic simulation model that is the target in this work needs to utilize a homogenized continuum material model at a macroscopic length scale to make it computationally efficient and robust.

An example of process induced material properties in different material orientations from an injection moulding material is presented in Figure 5. The polymer flow has been injection moulded from the right side to the left in the thin plate hence an orientation of the structure is horizontally in the MD-orientation. Dog bone shaped samples have been exposed to mechanical load and the force vs. displacement data, originating from Paper IV, shows a monotonically increasing force in MD and 45° material orientation. However, the force data in CD material orientation showed a decreasing force after the maximum peak has been reached once the onset of plastic deformation has occurred. Simulation model of the MD and CD material orientation is inserted as an example in Figure 5 to show the difference with respect to the localisation during the deformation where a localisation occurs in CD.

**Figure 5.** Experimental tensile test results in MD, CD and 45° for an injection moulded polymer material, extracted from Paper IV.
The “true” stress vs. strain relation, as presented in Paper IV, showed a monotonic stress increase, i.e. no material softening was observed at all. These phenomena highlight the difficulties when extracting the material constitutive behaviour from an experimental test and the assumption of a homogenous deformation in the specimen is not valid. These findings motivate local extraction of the deformation/strain field information to create a material model of the non-homogenous deformation in the specimen. In this study, the complete behaviour of the continuum mechanical response i.e. both the s-shaped curvature and the later part in the mechanical test, i.e. damage initiation and damage propagation, is of interest. The reason is that the fracture process also needs to be included in the FE-model describing the package opening sequence. An example from an injection moulded polymer is illustrated in Figure 6, illustrating this s-shape in Figure 6b). Injection moulded polymer tops, caps and opening devices utilize this type of information in the material model in the virtual opening simulation. Polymer films produced with extrusion coating show similar behaviour.

Figure 6a) present the force vs. displacement data acquired during experimental tensile test of an injection moulded polymer material and the inserted simulation pictures A-E shows the corresponding deformation of the specimen at different stages during the deformation process. The experimental and simulated curves in Figure 6a) show similar behaviour. Moreover, the associated and calibrated constitutive material behaviour utilized in the simulation model is shown in Figure 6b). The deformed simulation model, A-E, shows that different regions of the specimen is located at different positions in this stress strain curve indicated by different colours, i.e. non-uniform deformation in the vicinity of the neck. This material constitutive relationship, i.e. the curvature in the graph is based on the local stresses and strains developed in the smallest and thinnest cross section during the tensile test. These stress and strain quantities, compensated and accounted for the area change of the specimen cross sectional area in the neck, describe the local strain field present in the localised region. These strain and stress quantities will later be referred as true stress (Cauchy stress) and true strain (logarithmic strain) in this work to highlight that this is the local measures that is presented. The reason to extract this information from experiments is that the constitutive material behaviour is based upon these quantities and in one of the general-purpose finite element software is this stress denominated Cauchy stress. Therefore, a good understanding and description of the complete material response curve serves as input to the numerical material model.
In this study all materials are treated as homogenous materials through the whole thickness with in-plane anisotropic mechanical behaviour. Out of plane properties has not been concerned in this study nor shear loading scenarios as described by Islam (2018). However, both aluminium foil and thin polymer layers are not necessarily homogenous through the thickness and there is evidence that this simplification must be modified in the future, as shown by Nordgren et al. (2012), Andreasson et al. (2013), Björn (2018) and Wahlström (2018). During the last decades fundamental knowledge, and progress of experimental techniques have been developed rapidly. In-line or in-situ measurements are possible and available at large scale facilities, for instance synchrotron facilities. When this detailed information is more reliable and accessible the corresponding updates of the current FE-models are possible to even more replicate the reality and account for all the process induced microstructure and properties.

As concluded in Paper III, a simple slip-line theory is not sufficient to describe the deformation mechanisms in the semi crystalline polymer films in this study. The initial slope in Figure 7 for a extrusion coated polymer film, i.e. stiffness, is gradually decreased during the deformation due to morphological changes, slip mechanisms, and eventually the slope increases and finally the strain hardening effect is noticeable. The deformation mechanisms occurring in semi-crystalline polymers are further described by Schrauwen et al. (2004). Similarly, as the injection moulded specimen described in Figure 6b) has extrusion coated polymer films, presented in Figure 7, a significant strain hardening i.e. a steep slope at large strains.
The simulation models developed in this work need to accurately predict the mechanical continuum material behaviour beyond the experimentally measured data to account for the localisation and micromechanical event ongoing during the latter part of the tensile test curve. Mahnken et al. (2014) have seen similar strain hardening behaviour due to alignment of polymer chains. Furthermore, to describe the material failure the continuum data has to be accompanied with the onset of damage initiation and progressive crack path, i.e. the damage evolution. This information needs to be quantified from experimental tests and implemented in the simulation environment. In this work the damage and subsequent failure of the corresponding material layer has been included as a continuum material damage approach. The stress-strain curve for a general material undergoing damage is described and summarized in Figure 8, cf. Abaqus (2019) and utilized in Andreasson et al. (2012) and Andreasson et al. (2013c). A damage variable $D$ is introduced to represent the degradation of the material presented in Equation (1). Onset of damage is indicated when $D=0$ and a fully degraded material stiffness is represented when $D=1$. In the context of an elastic-plastic material with isotropic hardening, the damage has two forms: softening of the yield stress and decrease of the elasticity.
The solid curve shown in Figure 8 above represents the damaged stress-strain response, while the dashed curve shows the continuum response if no damage occurs.

\[ \sigma = \bar{\sigma} - D \bar{\sigma} = \bar{\sigma} (1 - D) \]  

(1)

The mechanical response at severe loading conditions leading to failure of the studied materials, have been further described in Papers I, III and IV. These methods have also been applied in Nordgren et al. (2012), Jönsson et al. (2013), Jin (2018) and Wahlström (2108). Similar mechanical behaviour, necking and constitutive relationship result in a polymer material were noticed by Diehl (2007). Moreover, this information was transferred and used as input to virtual material models for prediction of the package opening sequence in Paper V. One of the aims is that the presented experimental and numerical approaches should be utilized to create an efficient and useful tool for decision support. In the future it will be possible to drive the package and opening device development with realistic and predictive simulation models to speed up the development time.

The results presented in the following sections and in the included papers define the experimental test procedure and simulation strategy used. The workflow can be beneficial and applicable in other areas and industries. Furthermore, the results can be transferred to a variety of applications. The philosophies and strategies can be
used for a different set of material combinations. Such applications could be flexible/stretchable electronics, car industry, mobile industry and the medical device industry where laminated structures are well represented.

Thin laminated metal foils on polymer substrate have recently been studied by Shahmardani (2018) and Bolzon (2012, 2017) and earlier by Kao-Walter (2002, 2004, 2011), Li (2006, 2007, 2011), Suo (2005), and Hutchinson (2014). In all these prior works, localisation and thinning of the metal foil is also noticeable. The polymer layers, attached with a sufficient adhesion level, suppress this localisation in the metal foil. This material interaction is important to further investigate and better understand. Moreover, this synergy effect is dependent on mechanical properties of the substrate and bond strength between the layers. Many industrial applications have locally severe or extensive loading far beyond the globally measured quantities, hence an extensive knowledge about the true material mechanics is needed and the underlying physics and mechanisms behind. Transfer and an “extended” material behaviour originating from experimental observations are important to extract. This correct material description accounting for the geometrical effects and locally high strains enables the model to accurately capture the correct material physics in the defined constitutive equations. This has been done both for the polymer and aluminium material layers and is implemented in Papers I, IV and V. Holger Aretz et al. (2014) present a similar strategy, extending the experimentally measured data to account for the multiaxial stress state occurring in the application and the local material behaviour, for thicker aluminium sheets used primarily in the car industry. Mao et al. (2009) used a methodology of experimental data extension when conducting aluminium bottle forming with a virtual simulation model. Polymer components are used in many industries, experiments like the ones explained in Figure 5 is utilized in the furniture industry as well, cf. Chen (2015). A comprehensive introduction to polymer mechanics, what experiments to perform and a specific focus on available material models is given in Bergström (2015).
4. Material Micro Mechanics

Micro-mechanical understanding of the local effects involved when a laminated material composed of polymer and aluminium-foil layers is exposed to mechanical load is of interest here. Cross-sections with developing bands of localised straining for the described laminate are shown in Figure 9. In the band of the plastic deformation slip-lines occur where the crystallographic planes are exerted to shear loading along planes that form a 45° angle to the specimen surfaces. A continuous change of the position of the slip-planes while the deformation is increased leads to the three different stages, (a) - (c), and geometries shown in Figure 9. The bond strength between the two material layers is assumed to be sufficiently strong, hence the interface remains intact in the homogeneously deformed areas of the aluminium foil, but not strong enough to prevail in the region between point A and B indicated in Figure 9. The stiffer layer deforms independently of the behaviour of the weaker layer. The localised plastic deformation in the stiff layer introduces large strains in the weak layer that forces the polymer to the large local deformations. The theoretical model, derived in Paper III, has been used to compute the fracture process assumes elastic-plastic von Mise’s material models in both materials.

Figure 9. Specimen cross-sections and slip-lines of the laminate during deformation; (a)–(c) show the localisation during the deformation and fracture process. The materials are stretched in y-direction (plane strain) cf. Paper III.
The findings, previously shown in Figure 6 and Figure 7 have been incorporated in the present studies to more realistically capture the deformation and fracture process in the polymer layer. Local neck formation and a substantial strain hardening is essential to include. Accounting for the evolution of the microstructure during deformation and slip mechanisms will further improve the theoretical model and prediction of how a laminated polymer film reacts during the deformation process when exerted to mechanical load.

It was also shown that the local plasticity leads to a decreasing and eventually vanishing cross-section ahead of the crack tip for both the laminate and their single constituent layers. Experimental results are examined and analysed using a slip-line theory to derive the work of failure. An accurate prediction was made for the aluminium foil and for the laminate but not for the freestanding polymer film. The reason seems to be that the polymer material switches to non-localised plastic deformation with significant strain-hardening. This phenomenon is for instance illustrated in previous section in Figure 7, an extended explanation is given in Andresson (2015), as well as in Papers III and IV. The force per unit length along the x-direction, perpendicular to the cross sections shown in Figure 9, becomes,

\[
F(\delta) = \begin{cases} 
  F = \frac{2}{\sqrt{3}} (\sigma_{BA}h_A + \sigma_{BL}h_L - (\sigma_{BA} + \sigma_{BL})\delta) & \text{for } \delta < h_A, \\
  F = \frac{2}{\sqrt{3}} \sigma_{BL}(h_L - \delta) & \text{for } h_A \leq \delta < h_L, \\
  F = 0 & \text{for } h_L \leq \delta. 
\end{cases}
\]

where \(\sigma_{BA}\) and \(\sigma_{BL}\) are yield stresses in Al-foil respectively the polymer film. Further details of how these equations are derived is described in Paper III. The adopted slip-line theory with a final inclination (1:2), as indicated in Figure 9, of the failed cross-sections was verified for both freestanding aluminium foil and laminated aluminium foil by inspection of SEM micrographs of failed experimental specimens in Paper III.

An in-plane mechanical and fracture mechanical characterization was done for the polymer film and aluminium foil layers in Papers I, III and IV. Furthermore, an equation that calculates the work of fracture was derived. This created a well-defined understanding of the mechanical performance of each material layer. Fracture mechanics theory was used to derive an analytical expression for prediction of the critical load for centre-cracked specimens. This equation is accurate both for freestanding aluminium foil and a laminate consisting of one-side laminated aluminium foil with a polymer material layer. This expression can be used when the
plastic region is much smaller than the crack length. The following relation, readily computed from evolving geometry of the cross-section, gives the critical value of the J-integral, $J_f$, cf. Broberg (1999),

$$J_f = \frac{1}{h_A + h_L} \int_0^{h_A+h_L} F(\delta) d\delta = \frac{1}{\sqrt{3}} \left( \frac{\sigma_{BA} h_A^2 h_L + \sigma_{BL} h_L^2}{h_A + h_L} \right)$$  \hspace{1cm} (3)

Experimental evidence of the fracture process ongoing in thin material layers at a micrometre length scale has been described by Suo et al. (2005), Li et al. (2006, 2007, 2011) and Mehmood et al. (2014). In this study several cross sections of both free standing and laminated materials indicated that plasticity governs the fracture process in the aluminium foil, visualized with SEM micrographs in Paper III. Moreover, the polymer will localise and strain harden in the vicinity of the failed aluminium foil layer when exposed to loading in a laminated scenario with an aluminium foil with sufficiently high adhesion level as shown in Figure 10.

The fracture or failure process of the freestanding aluminium is a localised plastic deformation and thinning until the cross section vanishes. On the contrary in the freestanding polymer, the localised plastic deformation was not observed. Instead plastic deformation occurs in diffuse regions that surround the crack tip. The plastic region increases to incorporate most of the test specimen at larger loads.

In the laminate, the aluminium layer behaved similarly to a freestanding layer, but the polymer layer switched to localised deformation seemingly forced to do so, to compile with the deformation of the aluminium, shown in Figure 10(b)-d).

![specimen cross-sections](image)

Figure 10. Specimen cross-sections during the fracture process a) free standing aluminium foil and b-d) describe a laminated polymer film, show the localisation during the deformation and fracture process. The materials are stretched in horizontal direction. Inspiration from Li and Suo (2006)
On the contrary, injection moulded polymers can show a propagating neck phenomenon when dog bone shaped specimens are exposed to mechanical load. Therefore, this local information can be captured by visual methods such as image analysis. Applying a speckle pattern onto the sample, enables tracking of the local deformation field, to make it possible to quantify and extract the true intrinsic material behaviour, as mentioned in Nilsson (2017). This local information is used as input to the virtual material description. The stress vs. strain results in Figure 11 shows two curves denoted a) and b) with identical values in the initial part of the curvature. However, at larger strains there is some discrepancy in the latter part, in the strain hardening region. It is believed that the reason is that the DIC-speckle pattern analysed with the computer software GOM Correlate (2019) have higher local spatial resolution and hence can capture the highest and most critical local deformation region much more reliable and accurate. Bolzon et al. (2017) has applied DIC methodology for thin sheets of Al-foils that are very sensitive and challenging to handle. It has been concluded that accurate local information, i.e. intrinsic material behaviour, from experimental tests is fundamental to be able to create realistic descriptions of the mechanical response in the virtual models.

![Figure 11. Experimental results from uniaxial tensile test of an injection moulded polymer material and extracted with a Speckle pattern and four-point measurement acquisition pattern, cf. Paper IV.](image)

**Figure 11.** Experimental results from uniaxial tensile test of an injection moulded polymer material and extracted with a Speckle pattern and four-point measurement acquisition pattern, cf. Paper IV.
Using an analytical expression, in this case Ramberg-Osgood, to parameterise and extending the mechanical data points acquired during experimental measurements i.e. the constitutive material behaviour, is critical to create a realistic physical material behaviour for the thin aluminium foil, shown in Figure 12 and described in Paper I. Simulation on the micrometre length scale has made it evident what is related to material constitutive behaviour, microstructure and related to topographical effects. So far, the microstructure i.e. grain structure and grain boundary has not been included in the model. This approach needs more detailed experimental evidence and measurements and simple to use numerical models, currently is Crystal Plasticity (CP) modelling available described by for instance Roters (2010) and Melbin (2016). The mechanical material behaviour and the surface geometry i.e. the topography of the cross section, cf. Larsson (2017), are combined in the micro-mechanical models described in Paper I and below. The geometry described in Figure 13 and additional SEM pictures of cross section of aluminium foil was utilized as input for the creation of the geometry of the aluminium foil cross section in the micro-mechanical model. SEM images are useful to increase the understanding and to visualize the highly local effects at high magnification and resolution. The mesh density contains several elements through the thickness of the aluminium foil and can readily capture the local effects ongoing through the thickness previously described in Section 4. The cross section is loaded mechanically in the horizontal direction.
A visualization of the cropped simulation model result, zoomed within the same limited area of the cross section, with the corresponding force displacement graph of the cross-section evolution, through the thickness, is presented in Figure 14. The deformation sequence in Figure 14b) is denoted I) to VII) where I) represents the unloaded initial geometry and VII) is the final geometry when the aluminium foil has failed. These numerical results can be compared with the experimental observations and outputs from SEM made in Paper III. The geometry is localized when the force reaches its maximum as shown in Figure 14b), between the two different load steps IV) and V). However, limited knowledge and information is acquired about the microstructure i.e. the local information about the grains, e.g. size, shape and distribution, and grain boundaries in the aluminium foil. Therefore, this lack of information has been identified as a gap and needs to be further addressed to be able to connect the microstructure with the mechanical performance.
Micro-mechanical simulations of the micro-mechanisms are a useful tool to understand the deformation mechanisms. This is a complement for increased understanding of the theories and hypothesis described in Paper III.

Understand the basic mechanics and transfer this information to analytical expressions and parametrized models to be able to virtually generate a set of mechanical material behaviour at the hand of the simulation engineer ready to utilize the information has enabled a standardized way of collecting experimental results and to make the information accessible to others.

Moreover, the mechanics, continuum mechanics and damage mechanics of thin layers of both polymers and aluminium foil have been the primarily focus in this work. The next sections describe how the simulation models have been developed by using the experimental results, deformation mechanisms and micro-mechanical observations at multiple length scales. Aluminium and polymer materials originates from different chemical compositions, microstructures and manufacturing processes. Therefore, the implementation may need to be treated in separate ways depending on the material layer. Both materials studied here are exposed to severe mechanical loading involving progressive damage.
Additionally, the aluminium foil has also a considerable difference in the global vs. local material response and a large increase in the strain and stresses in the localization where slip-line formation occurs due to the microstructure composed of different grains with different sizes, shapes and local orientation. Therefore, this work has pointed out that it is important to combine different experimental techniques and to be able to visualize or record the deformation sequence with a camera. This in order to combine and correlate the geometrical deformation and mechanisms from pictures with the acquired signals of deformation and force recorded by the measurement equipment.
5. Experimental and Simulation Strategy

Experimental testing is often easy and relatively straightforward to setup and to perform. However, it can be a long and tedious process interpreting, evaluating and analysing the acquired test data described further in Papers I-IV. An experimental test strategy has continuously been developed and refined during this research project as shown in Figure 15. The experiments range from uniaxial tensile test, cutting resistance, trouser tear test, peel test and finally opening torque. An inverse analysis procedure, i.e. performing the same experimental test in the virtual environment, has been utilized to identify and quantify the material model parameters utilized in the simulation model with an optimization scheme. In-plane mechanical properties are the governing quantities that were determined. The trouser tear testing, presented in Andreasson et al. (2013), showed that in a highly extensible polymer film it is difficult to separate the leg extension, the plastic flow and the actual tearing force during the result evaluation. The plastic flow at the crack tip is not solely involved in the fracture process and hence the deformation does not only take place locally close to the crack tip. Therefore, it is hard to find solely one material parameter governing tearing in this type of studied material. The governing material effects were experimentally identified, observed, understood and characterized, e.g. in the applications.

![Physical and Virtual Tests combined with inverse analysis to calibrate the parameters.](image)

Figure 15. Physical and Virtual Tests combined with inverse analysis to calibrate the parameters.
The next step was to include the major contributing effects in a step-wise approach and account for them one by one in the virtual test replicating the experimental test. Accounting for morphology and microstructure in the studied materials are challenging due to the small length scales involved compared to the macroscopic length scales that is modelled with the existing hardware. Therefore, the microstructure is not explicitly modelled in the simulation models. A homogenization of the continuum material and a smeared continuum damage modeling approach on a mesoscopic length-scale are utilized to include and account for the multiple length scales involved. The evolution of the microstructure that is ongoing during the deformation process is therefore moved and homogenized to a higher length scale compared to the length scales at which the microstructure is changed to be able to solve the application at hand with the correct physics involved. In the general purpose finite element software, the material model, i.e. constitutive relationship, is defined with true stress and logarithmic strain quantities as previously shown for the mechanical response in Figure 6 and Figure 7. The overall simulation strategy is further explained and described in more detail in earlier work by Dabiri et al. (2012), Nordgren et al. (2012), Jönsson et al. (2013), Nordlund et al. (2014) and in Paper V. Pagani et al. (2012), Bolzon et al. (2012, 2015, 2017, 2018) and Shahmardani (2018) have developed experimental and numerical strategies for the thin material layers used in the packaging industry.

Several approaches exist of how to numerically treat a material subjected to a critical loading condition including severe plastic deformation and progressive damage. Hierarchical material modeling approaches can span from a micro-mechanical to meso- or macroscopically nature. The decision of what model to select depends on the need of accuracy and on which application and questions that is aimed to be solved with the model and how this information will be incorporated into the simulation model. User subroutines written in Fortran are a common way of introducing customized, complex and flexible material models into commercial general-purpose finite element software’s, e.g. Abaqus™. A phenomenological description of the material mechanics was finally adopted with ordinary built-in functionalities in the finite element solver. The material models consist of a small number of material model parameters to make them feasible and easy to use in industry. However, independent of which material description that is used several repeatable, reproducible, accurate and reliable experimental tests are fundamental and needed in order to characterize the mechanical material behaviour in the studied
material. Material model parameters determining the continuum and damage criterion for the specific material needs to be calibrated and fine-tuned with this experimental data.

An explicit simulation scheme was chosen as the primarily tool in the opening application simulations, this is maybe not an obvious choice. However, contact algorithms are much more mature and easily adopted within this framework. Progressive fracture modeling is also a conditionally unstable event and is most often hard or even impossible to solve in an implicit code framework without utilizing several stabilization options and convergence tolerance reductions. Explicit codes are often utilized for rapid and dynamic events like a car crash or drop test of a mobile phone. The package opening process is, on the contrary, a slower event in comparison. Small elements resolve a high resolution but have an additional computational cost in an explicit code. Thus, decreasing the time increment extends the time to solve the numerical model if the total time event is rather long in reality. Different methods have to be utilized such as semi-automatic mass-scaling to find a good balance between the total simulation time and accuracy of the simulation results. Furthermore, the physics and material mechanics should never be violated.

Numerical material models have been created and calibrated, e.g. a virtual Packaging Material is today being used by the packaging industry. These simulation models capture the experimental observations and results and in-plane mechanical behaviour for polymer film and aluminium foil at different temperature and strain-rates exposed to different loads. In both the polymer film and aluminium foil, the governing material properties were found to be the in-plane anisotropic continuum mechanical material behaviour. Furthermore, the fracture process is governed by the microstructure, i.e. the orientation of the individual grains in the aluminium foil, the alloy composition and alloy elements that build up the grains. In the polymer film there is a combination of how the molecular chains are arranged and how the structural arrangement of the amorphous regions, crystalline regions and crystallites are mixed and organised, described by for instance van Dommelen (2004) and Björn (2018). Severe local plastic deformation are the governing fracture processes in both materials until the cross section vanishes, as concluded in Papers I and III.

The experimental and modeling strategy has stepwise been refined and extended to include more functionality. Experimental observations and mechanical behaviour are linked and incorporated in the simulation models. It is important to have an “end-to-end” perspective including all the necessary functionalities and components to build
an accurate, reliable and realistic opening simulation model. If something is changed in the packaging material, the manufacturing process or late design changes of the opening device, the simulation model should be able to capture these changes accurately to distinguish and predict the new opening performance response. Therefore, it is important to separate each individual packaging material layer and treat them as individual packaging material layers with respect to geometry and mechanical material properties. All these factors enable the utilization of a simulation model for simulation-driven design and to make information based on fact available to ease the decision support.

Paper V describes the final simulation opening model at the application level and puts the separate pieces together. It is concluded that it is possible to realistically predict the opening performance in an opening simulation with the finite element simulation model. The main advantage with the developed modelling approach is that the strategy is very flexible. Material layers can be altered, and the opening device can be changed in the model. The simulation results mimic the experimental results satisfactorily. Therefore, this type of simulation models can be used for decision support early in the concept selection phase. Simulation models where both the packaging material and the opening device are included in the model can potentially help in the development of packaging materials and the opening devices and act as a decision support tool for future development. Design variants of opening devices and different material combinations can be tested and evaluated in the virtual environment early prior to manufacturing. This methodology shortens the development time and enables a simulation driven development approach.

Furthermore, simulation models that should be used for predictive purposes in the daily work need to be very easy to use, simple to calibrate, efficient and fast to solve, accurate, stable and reliable. The far most important part is to understand the physics and different mechanical events occurring during the deformation process and to include and trigger these features in the realistic simulation model. If the mechanisms and phenomena are identified and mechanical response/behaviour is measured experimentally there is no need of handbooks and constants. Moreover, the simulation models can be widely utilized and re-used. The same standardized approach could be used through a whole organization and in the future, it is possible to transfer the complex “expert” simulation methods and models to “non-expert” users within the organization i.e. democratization of simulation models.
6. Virtual Package Laboratory (VPL) for Decision Support

The Virtual Package Laboratory (VPL) consist of virtual simulation models based on realistic beverage packages. These virtual package twins are utilized during package development, filling machine development and when root cause analysis is performed. Three components are instrumental in these functional virtual models; I) Realistic geometrical representation at the studied/modelled length scale i.e. a virtual geometrical configuration, II) Accurate and realistic constitutive material relations of the mechanical response of the packaging material i.e. a virtual packaging material and finally III) boundary conditions and load scenarios from realistic load cases.

A fundamental understanding of the mechanics involved during mechanical deformation of paperboard complemented with information described here of the thin polymer and aluminium-foil layers that a package is constructed of has enabled a VPL in the packaging industry. A continuous improvement and problem-solving methodology approach have been applied during this work. The modelling strategy has stepwise been refined and extended to include more functionality. Experimental observations and mechanical behaviour from experiments is linked and incorporated in the simulation models today. It is important to have an “end-to-end” perspective including all the necessary functionalities and components to build an accurate, reliable and realistic opening simulation model. The exclusion of the none-important or not activated features is equally important in the application. If something is changed in the packaging material, the manufacturing process or design changes of the opening device, the simulation model should be able to capture these changes accurately to distinguish and predict the new response at the application level. Therefore, as it has been mentioned before, it is important to separate each individual packaging material layer and treat them as individual packaging material layers with respect to geometry and mechanical material properties. All these factors enable the utilization of a simulation model for simulation-driven design and to make information based on fact instead of gut feeling available to ease the decision support.

Paper V, which describes the final opening simulation model at the application level and puts all the separate pieces together, shows that it is possible to realistically predict the opening performance in a virtual opening with the finite element model developed during this work. The advantage with the developed modelling approach is that the strategy is very flexible. The simulation results mimic the experimental
results satisfactorily. Therefore, this type of simulation models can be used for decision support early in the concept selection phase. In addition to this, multiple loading scenarios and applications can be studied nowadays. This is possible due to the fundamental understanding of the two packaging material layers in combination with the existing and available data of paperboard. Most of the developed building blocks have already successfully been implemented in the current package simulation workflow. Simulation models where both the packaging material and the opening device are included in the model can potentially assist in development of packaging materials and the opening devices and act as a decision support tool as shown in Figure 16. The simulation models are also an efficient tool to integrate and link the development of the two parallel development activities. Design variants of opening devices and different material combinations can be tested and evaluated in the virtual environment prior to manufacturing. This methodology shortens the development time, front-loads the work and enables a simulation driven development approach with a superior knowledge and understanding of the opening system developed. Furthermore, simulation models that should be used for predictive purposes in the daily work need to be very easy to use, simple to calibrate, fast to solve i.e. computational efficient, mature, accurate, stable and reliable.

Today’s simulation models are more often utilized early in the development phase and can therefore be proactively used as guidance and support during a whole project’s different phases. Everything from early technology development and idea screening to very late in projects when facing/solving issues and problems in the field. However, continuous improvement is always important and to be agile now when entering the digital era that is now emerging it is probably even more important to have reliable, sufficiently accurate and mature simulation capabilities at hand for projects to use. It is important to be on the forefront of technologies both on the virtual and experimental side.

Figure 16. Simulation model with decision support enables simulation driven development integrating packaging material and opening device criteria’s
7. Summary of the appended papers

Below is a short summary of the content and the respective main findings from the five included papers, Papers I-V:

**Paper I**

Experimental uniaxial tensile test results of the in-plane mechanical behaviour of thin aluminium foil in the principal material directions (RD, DD, TD) are presented in this paper. A mathematical expression is fitted with a direct method to the test data and a numerical material model is selected and calibrated to reproduce the mechanical response of the thin aluminium foil from the experiments in the simulation model. An extension of the experimental data with the analytical expression, derived by Ramberg-Osgood, enables a good description of the material mechanical behaviour in true stress vs. true (logarithmic) strain diagram i.e. the local material behaviour. Numerical material model parameters were determined.

**Paper II**

This paper is a mathematical postprocessing analysis of the initial part of the experimental tensile test data presented in Paper I. Performed in 10 different material orientations ranging from 0° to 90° to the rolling direction of the same thin aluminium foil as presented in Paper I. The plastic deformation initiates and evolve almost immediately at low strain levels in the beginning of the deformation giving a virtually vanishing yield stress. Furthermore, material parameters in the initial mechanical response i.e. the elastic modulus in all the different material orientations was determined and the corresponding principal material directions. This study shows that the elastic modulus seems to have almost similar values as bulk aluminium material despite the thin thickness of the aluminium foil.

**Paper III**

The fracture process is investigated and discussed in both thin extrusion coated polymer film and aluminium foil exposed to mechanical loading. Freestanding layers and the corresponding laminate are considered. Visual inspection of the broken cross sections from the experiments with the aid of SEM is performed to reveal the material mechanics and micro-mechanisms ongoing in the vicinity of the failure. Experimental results are examined and analysed using slip-line theory.
**Paper IV**

In-plane anisotropic mechanical material properties of an injection moulded polymer LDPE-plate have been investigated. Uniaxial tensile tests were performed, accounting for the developed local strain field on the surface of the test specimens using image analysis. Three different strain rates have been included in this study. “True” stress vs. logarithmic strain diagrams were constructed based on the experimental results. Furthermore, these mechanical responses were implemented and evaluated using finite element simulations. Numerical material model parameters were determined.

**Paper V**

In this final paper the conclusions from previous works and related work are implemented and the opening device application is simulated in a general-purpose finite element software. The results presented in this paper show that it is possible to select numerical material models with corresponding continuum material damage descriptions adequately predicting the mechanical behaviour and failure of thin laminated packaging materials. Finally, comparison between the experimental opening and the virtual opening results, showed a good correlation with the developed finite element modelling technique.

The five included papers summarized above in this PhD-thesis describe three different building blocks needed to create an accurate simulation model describing the application presented in Paper V. The building blocks consist of (i) *experimental mechanics and visualization techniques*, (ii) *material science* i.e. manufacturing processes, material structure, continuum mechanics, constitutive modelling, micromechanisms, damage and fracture processes and finally the gained and build knowledge implemented in the third building block (iii) *simulation modelling techniques and strategy*. A combination and good balance of these three building blocks results in an engineering sound and realistic finite element modeling strategy. A bottom up approach, i.e. understanding the fundamental components that a packaging material consists of, as described in this work, may take a long time to create. However, the advantage and benefit are when this work is finalized it’s possible to predict realistic behaviour in the simulation models and utilize the same information in a multitude of loading scenarios and different applications. Therefore, the virtual twins can be utilized to support decisions during the different tollgates within projects in the industry. The objective within this work has been focused on
the development of package opening simulations available in the packaging industry that eventually will be used for decision support.

In the present research work the focus has been on thin material layers, composed of polymer (LDPE) and metal (aluminium foil) layers, present in a food beverage package. Several experimental mechanical tests have been performed to characterize the mechanics and fracture process involved during mechanical loading of single and laminated packaging materials. In-plane continuum material response, Papers I, III and IV, and out-of-plane mechanical behaviour are triggered in the performed tensile and trouser tear tests, Andreasson et al. (2013a). The methods discussed will help to quantify and classify different groups of packaging materials. The polymer film studied is oriented and extensible. In-plane material orientation/alignment induced during manufacturing creates anisotropic in-plane mechanical properties, Papers I and IV.

When the packaging material membrane is stretched in the initial phase of the cutting sequence the membrane response is dominated by the in-plane mechanical properties. Subsequently during the next phase of the opening, i.e. the cutting of the laminated membrane, the trouser tear testing is a good complementary experimental test to characterize the fracture process and fracture path of different polymer materials, further described in the related work. A brittle polymer film is dominated by Fracture Mode III and for a ductile polymer film the fracture process is a mix between Mode I and Mode II. The locally stretched polymer material is involving a much larger region outside the vicinity of the crack. Therefore, the in-plane mechanical material properties come into account and local high elongation and plastic deformation is observed in the laminated tensile test specimens, Paper III. The experimental information presented in Papers I-IV is important and necessary to understand to be able to simulate at the application length scale, presented in Paper V. A multiple-length-scale approach was used throughout this work, as indicated in the different length scales that are typically involved in an opening application, shown in Figure 17. In the figure the opening application is illustrated to the left and the circular membrane that is cut during the opening sequence is in the middle of the figure. To the right in Figure 17, is a SEM-micrograph of one of the polymer layers present in the material structure. Papers I-IV highlights the benefit of looking at multiple length scales and especially at the smaller length scales, ranging from nm-mm. The gained information is fundamental to base the decision of how to numerically model the materials and what effects and physics to include in a
simulation model. Moreover, it is possible to identify and characterize the involved deformation mechanism in the local thus microscopic behaviour that is most often measured at a macroscopic level in the experimental test.

Geometrical effects such as necking, thinning and structural changes are involved in the tensile testing and accounting of these effects is needed to be able to extract the true intrinsic mechanical material behaviour from experimental tests. The true mechanical material behaviour is later implemented and used in the numerical material model parameters to be able to represent the polymer film and aluminium foil realistically with correct physics in the model.

In Paper V, the experimental and numerical strategies are combined and connected to create a baseline and hence a nominal application simulation model for one specific package opening device. To verify the fidelity of the selected approach it is necessary to link experiments with virtual tests. Furthermore, the different building blocks are combined, i.e. the numerical implementation of the mechanical behaviour of individual material layers in a commercially available general-purpose finite element solver, Abaqus™. The experimentally identified deformation mechanisms and effects are considered. A numerical strategy is created and finally a simulation model is developed and proposed. The proposed model is appropriate to solve industrial applications on a macroscopic length scale with an efficient and affordable computer model.

![Diagram](image.png)

**Figure 17.** A beverage package with a post applied opening device, with the corresponding material structure at different length scales involved in the application presented in Paper V.
The focus and maturity level reached within different disciplines in the different research papers, Papers I – V, is shown in Figure 18. The ambition to create a “Simulation Driven Design” during the package opening development has been the target and long-term goal in the performed work. Mechanical Behaviour and Deformation Mechanisms are inherited and linked to the Manufacturing Process, therefore these disciplines have been studied in most of the Papers and reached a high maturity. Experimental Techniques, Simulation Strategy and Predictive Models have been included in the Papers. However, Material Interfaces have been included in the work but needs more attention in the future. Decision Support has been included in all Papers and has gradually been improved and has finally reached a high maturity. Today it is possible to base the decisions solely on mature simulation models during the design of new opening concepts and opening designs. This increase in maturity is the result of the build fundamental understanding and good description of the material mechanics that is now available.

The papers are organized according to the following structure; Papers I-IV focus on experimental techniques, visualization of the micro-mechanics, process induced properties, experimental results and analysis and extraction of true intrinsic material
behaviour based on experimental data available from individual packaging material layers. The work has considered both polymer and metal materials. Paper V describe and presents the modelling strategy and how the gained knowledge from Papers I-IV and the related work is introduced and utilized in the creation of the simulation model of the opening sequence.
8. Conclusions

Below is a summary of the highlights and conclusions made during and within the framework of this PhD-thesis:

- A user-friendly finite element modelling strategy targeting an industrial available package opening application has been developed.
- It is possible to predict the opening performance with realistic results in an opening simulation with the developed finite element simulation model. Comparison of the experimental opening results with the corresponding virtual opening model, shows a good correlation.
- The governing deformation mechanisms and mechanical properties involved in and during the opening sequence were quantified with full field experimental techniques.
- “True” stress vs. “true” strain diagrams based on experimental data were possible to construct for the injection moulded polymer, extrusion coated polymer and aluminium foil material.
- An identification process to calibrate the suitable numerical material model parameters with inverse modelling analysis is recommended.
- The in-plane continuum mechanics and fracture process of extrusion coated and injection moulded polymer materials is possible to accurately simulate in the virtual simulation model.
- Aluminium foil has been thoroughly studied and simulated both at a micro- and macroscopic length scales with good predictions of experimental results. Both in comparison to the mechanical response and to the corresponding localisation visible in the cross-sectional view of the broken specimen.
- The fracture or failure process of the freestanding aluminium is a localised plastic deformation and thinning until the cross section vanishes.

The main effort in this PhD-thesis has been on material characterization of thin layers of aluminium foil and polymer materials. Mechanical, structural and surface topographical information has been investigated to increase the knowledge of these two materials studied. Experimental methods combined with virtual methods and tools have continuously been developed. Today, the awareness, maturity and knowledge are good in respect of different length-scales involved, microstructure and geometrical features present in the studied materials. Mechanical response during loading is implemented in the simulation models and the results are replicated by the
virtual twins, including non-linear effects, geometrical changes, material orientation and progressive damage modeling. Experimental test-data is linked to numerical descriptions of the mechanical response in the two different materials.

Experimental observations at different length scales has served as input to the simulation models and resulted in an increased understanding of the material mechanics and deformation mechanism of single material layers and laminated packaging material layers. In addition to this, the experimental results were used as verification and validation of the simulation models. It has been found that the fracture process of the freestanding aluminium is a localised plastic deformation and thinning until the cross section vanishes. On the contrary a considerable strain hardening is occurring in the semi-crystalline polymers, when these material layers are localised during mechanical load. These polymer materials where manufactured with injection moulding or extrusion coating.

Furthermore, several experimental techniques have evolved, improved and been simplified during this work. For example, DIC technique has lately been possible to study with an ordinary system camera with high resolution and combined with a software, GOM/Correlate, available from the German supplier GOM. Extraction and creation of numerical material models solely based on experimental data points, i.e. force displacement, is not enough. The acquired data combined with the local information i.e. intrinsic true stress/strain obtained with speckle pattern or similar methodology is required. This is valid for both the studied polymer materials manufactured with extrusion coating technology or injection moulding technique. Capturing the significant strain hardening present in the polymer during mechanical deformation as pointed out in Paper III in the propagating neck region is critical to represent the constitutive relationship in the material at large strains.

This PhD-thesis present a methodology of how to extract the constitutive material behaviour based on experiments in semi-crystalline polymers manufactured with injection moulding and extrusion coating technology. Moreover, the mechanical response in thin aluminium foil can be accurately described with an extended constitutive description, including the hardening present in the localisation during tensile tests, based on the analytical Ramberg-Osgood equation. These two different approaches and fundamental material information can then further be utilized to implement and calibrate a numerical material model in a general-purpose finite-element software. The determined material mechanics description can be utilized in a wide range of applications where for instance opening simulations are an example.
where the complete material curve, i.e. mechanical behaviour with the associated microstructure deformation, damage initiation and the fracture processes is included. The developed methods are not only restricted to be utilized within the packaging industry. On the contrary, the described methods and tools are general and can be adopted and implemented in other industries. Thin and thicker polymer material layers have been considered in this work. Therefore, extracting the true intrinsic mechanical response, i.e. local material behaviour, is the fundamental start in all accurate material modelling activities and serves as a pillar and starting point in material modelling. Thus, creating a virtual material model that will be utilized when solving applications with the aid of simulation models.

The deformation process and involved phenomena and mechanisms occurring during the opening of a beverage package are rather complex to simulate. Therefore, the work herein has been divided in a stepwise approach to understand the major contributions. A multiscale approach is useful reaching from micro-mechanical observations and simulations to the application level at both meso- and macroscopic length scale. Reliable and calibrated in-plane numerical material models describing the mechanical behaviour in thin layers of packaging materials have been successfully developed during this work. Suitable constitutive models for the continuum material and how to address the progressive damage modeling have been presented. The inverse modeling technique combined with video recording of the involved deformation mechanisms in the experimental tests was utilized for identification and calibration of the numerical material model parameters. Non-linear anisotropic material behaviour with significant strain hardening at large deformation, bond strength and fracture are all identified effects that need to be included in the virtual opening model. These described methods can be adapted to other industries where semi-crystalline polymers, thin aluminium materials and laminated materials are present. Therefore, these studies can be utilized in a wide range of applications and simulation models.

The numerical results are compared with experimental data at macroscopic length scales in an application. A good prediction of the opening force and overall behaviour was achieved at nominal settings of all the involved parameters, shown in Figure 19, and furthermore detailed described in Paper V. The simulation model includes the correct physics and material mechanics. Each packaging material layer is modelled as a homogenized in-plane anisotropic material model description. Moreover, the
deformation mechanisms with large deformation and material interfaces are included in the simulation model.

The new knowledge and findings have in parallel to the ongoing research activities been transferred and adapted in the packaging industry to solve industrial engineering problems. The presented applied research work has been a joint effort and a good collaboration with different partners depending on topic. Limited knowledge about the different packaging materials were available initially and this fundamental knowledge building have enabled the Virtual Package Laboratory.

These presented results available today was almost a fiction and utopia when this thesis work was initiated, but today it has become a reality.

Figure 19. Comparison of the experimental and virtual opening tests cf. Paper V.
9. Future Work

The continuation of this work is focused on implementation of several of the explored techniques studied and utilized during this PhD-work in the packaging industry. For instance, implement full field methods such as DIC. Systematic fracture mechanical testing and quantification of material quantities associated, and governing damage initiation and propagation need to be studied. Cutting resistance, i.e. cutting principles, and adhesion with corresponding interface properties needs enhanced knowledge to further understand and implement in the simulation models. Furthermore, to highlight the possibility to include and use progressive damage modeling more easily as an engineering tool for the simulation engineers in their daily work.

Process induced material properties due to the manufacturing principal, geometrical tolerances and noise factors has to be quantified and accounted for in the next step of simulation models. It is important to create an easy-open and robust design of the beverage opening process. The focus has gradually shifted to improve and include the geometry used as input of the simulated components as the material mechanics have matured. Computerized Tomography (CT) scans complemented with high resolution SEM-pictures is and will be a very useful tool for these purposes. Postlind et al. (2016) has for instance done “static” peel tests loaded within a CT-equipment. The next step is to be able to acquire information continuously during loading. Experimental techniques have the latest years moved the forefront and the possibilities and techniques are much more available and accessible today. Large scale facilities enable these high-end experimental techniques and the acquisition is much faster compared to laboratory equipment. Hence, in-situ loading and experiment to visualize in a realistic time frame is almost possible. Opportunistic will we soon be able to, in a CT-environment, to open a package in a realistic manner. Hence experimental methods and virtual methods will always go hand in hand and soon it is possible to do it simultaneously at the same order of time and length scale to be able to compare the both results directly to understand the reliability of the models.

As stated in the introduction section the material behaviour and the underlying deformation mechanism (geometrical and structural) is challenging to identify, despite substantial material testing. The local deformation and hence the local strain field cannot be captured in the overall macroscopic measurements that are performed.
in experiments today. Therefore, Digital Image Correlation (DIC), in-situ SEM testing, X-ray computerized tomography (XCT), in-situ XCT or similar advanced techniques with the full field capabilities are needed in the future to better understand and account for the local and microscopic deformation mechanisms occurring in the experimental setup during the experiments. These techniques enhance the possibilities to identify and extract the “true” local mechanical material behaviour. Combinations of complementary and supplementary techniques are very important to observe and extract more useful information for enhancing the simulation models. In the future, the new large-scale research facilities currently being built in Lund, MAX IV and ESS, will further enhance these capabilities. In-situ testing, imaging during deformation at operating conditions is also a key to increased understanding and knowledge of the packaging materials and the opening sequence. Digital volume Correlation (DVC) described by Tudisco et al. (2017) can also be an option in the near future.

Direct methods complemented with full field techniques or even volumetric capturing techniques available with high spatial resolution with fast acquisition at large scale facilities for instance MAX IV or ESS shown in Figure 20 below will be important tools for an increased material understanding.

Figure 20. MAX IV and ESS, examples of two large-scale facilities soon available in Lund, Sweden that will enable much more detailed material science studies, based on X-ray or neutrons with e.g. high resolution and rapid time resolution, courtesy of MAX IV and ESS.
The large deformations involved when ductile polymers are modelled are typically difficult to handle in a reliable way with a Lagrangian element formulation often used in a FE-model. In this work primarily, shell elements with reduced integration were utilized. An Arbitrary Lagrangian Eulerian (ALE) approach or continuous remeshing technology complemented with triangular elements that does not suffer to a similar extent when exerted to large deformation would be more appropriate. The quadratic element shape that was used suffers from increased aspect ratio directly when the specimen is stretched during the tensile test. The eXtended Finite Element Method (XFEM) for instance available in the commercial finite element software RADIOSS™ or Abaqus™ could be an alternative to reduce the element shape and element size dependency in the crack initiation and propagation path. Continuum Damage Modelling (CDM) has several benefits in respect of simple to use. However, some deficiencies must be solved or investigated future on. For instance, the need of finding a reliable and simple strategy of how to extract the numerical material model parameters describing the onset of localisation/damage and the propagation i.e. damage evolution parameters from experiments. Element size and some local length scale parameter must be considered.

Virtual twins are successfully used in many industries today. Furthermore, the need of including more and smaller details in the FE-models is continuously pushed from the projects. One of the reasons is that the virtual package models are very accurate today and the awareness is sometimes very high, hence more can be accounted for and matching reality in the computer is the key to be able to predict realistic scenarios. Therefore, it is a necessity to further enhance the detailed knowledge about the different material components and the process induced properties and effects. The building blocks created during this work has to a large extent already been implemented and are utilized within the packaging industry and a more high-end and advanced experimental test protocol complemented with systematic fracture mechanical testing is the next hurdle to climb over. Advanced experimental techniques such as SAXS, WAXS, EBSD, FIB, XCT, DCT complemented with in-situ capability to name a few needs further attention to be able to quantify smaller length scales and to characterize, visualize the microstructure in the reality that should and could be fed into the future simulation models.
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Simulation of Thin Aluminium-foils in the Packaging Industry

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Abstract. This work presents an approach of how to account for the in-plane anisotropic mechanical material behaviour in the simulation models of the thin aluminium foil layer (\(\approx 10 \, \mu\text{m}\)) used in the Packaging Industry. Furthermore, the experimental results from uniaxial tensile tests are parameterised with an analytical expression and the slope of the hardening subsequently extended way beyond the experimental data points. This in order to accommodate the locally high stresses present in the experiments at the neck formation. An analytical expression, denominated Ramberg-Osgood, is used to describe the non-linear mechanical behaviour. Moreover, it is possible with a direct method to translate the experimental uniaxial tensile test results into useful numerical material model parameters in the general finite element software Abaqus\textsuperscript{TM}. In addition to this the extended material behaviour including the plastic flow i.e. hardening, valid after onset of localisation, the described procedure can also capture the true intrinsic material behaviour present at the microscopic events, i.e. geometrical thinning, ongoing in the deformation of the aluminium foil. This method has earlier been applied cf. Petri Mäkelä for paperboard material [1]. The engineering sound and parameterised description of the mechanical material behaviour facilitates an efficient categorisation of different aluminium foil alloys and aid the identification of the correct anisotropy (RD/TD/45\(^\circ\)) of the mechanical material behaviour based on the physical testing.

MECHANICAL CHARACTERISATION OF ALUMINIUM-FOIL

Thin sheets made of different metals are applied in multiple industries today such as flexible/stretchable electronics, solar panels and packaging industries to name a few. Several researchers have been focused on the mechanical properties and the damage process involved in such materials cf. [2]-[7]. In the packaging industry recent advancements in the finite element models have called attention of improved material models for thin aluminium foil. Therefore, an increased knowledge is needed of how the aluminium foil reacts to the prevailing loading scenario. This work focus on the experimental mechanical characterization of the thin aluminium foil, depicted in \textbf{FIGURE 1}. Thereafter, the information acquired from the experimental tests i.e. non-linear mechanical material behaviour, is transferred into numerical material model parameters with the aid of an analytical expression. The experimental part have been complemented with FEM-simulations replicating the experimental setup to further increase the understanding of the mechanical behaviour and the deformation and damage mechanisms involved on both macroscopic and microscopic length scales.

Käck et al. [8] investigated both an advanced and basic mechanical characterization of a thin aluminium foil. Furthermore there is limited knowledge and experience about the anisotropic material behaviour in general, the associated damage process and how the material properties in the smaller length scales actually influence the macro mechanical behaviour. Experimental tests on this specific aluminium foil have earlier been limited to uniaxial loading in the manufacturing direction i.e. \textit{Rolling Direction (RD)}. A numerical material model that reflects the anisotropic behaviour observed in the experiments is therefore needed. The purpose of this work is to thoroughly characterize aluminium foil and the work is meant to be used for future studies of aluminium foils.
Experimental measurements were performed on rectangular samples with dimensions 15 x 100 mm in accordance with the ISO Standard 6892-1 [9] at controlled climate, i.e. 23°C and 50% relative humidity. The experimental result with extracted material parameters from the basic mechanical characterization is according to FIGURE 1, TABLE 1 and FIGURE 2 [8]. During this project, one specific alloy, AA8079, and type of aluminium foil with a given thickness has been studied, implying that the results may vary for other foils. The material has been treated as load symmetric i.e. tension and compression shows identical mechanical material behaviour. Furthermore, an extension of basic characterization was done with an incremental angle of 10 degrees to RD referred as advanced material characterization, described in [8]. The material parameters determined and used in this study needs to be validated with a biaxial or a multiaxial test such as a bulge test. Three material directions were used for the model.

### TABLE 1. Mechanical material parameters used as input to the numerical material model [8]

<table>
<thead>
<tr>
<th>Direction</th>
<th>angle α [°]</th>
<th>E [GPa]</th>
<th>σα [MPa]</th>
<th>σ_{uts} [MPa]</th>
<th>ε_{uts} [%]</th>
<th>W_f [J/mm³]</th>
</tr>
</thead>
<tbody>
<tr>
<td>RD</td>
<td>0</td>
<td>35.74 ± 5.41</td>
<td>40.66 ± 0.55</td>
<td>74.41</td>
<td>4.14 ± 0.09</td>
<td>2.55</td>
</tr>
<tr>
<td>TD</td>
<td>90</td>
<td>35.06 ± 3.50</td>
<td>38.23 ± 0.21</td>
<td>65.76</td>
<td>4.03 ± 0.38</td>
<td>2.14</td>
</tr>
<tr>
<td>45</td>
<td>45</td>
<td>32.67 ± 1.99</td>
<td>36.91 ± 0.30</td>
<td>67.96</td>
<td>7.36 ± 0.22</td>
<td>4.38</td>
</tr>
</tbody>
</table>

**FIGURE 1.** Mechanical characterization test results from in-plane mechanical material testing of thin Aluminium-foil [8].

**FIGURE 2.** Definition of material parameters extracted from the experimental tests [8].
Noticeable in FIGURE 1 is that the material directions RD and TD is extended to a similar maximum displacement, on the contrary is the maximum displacement in the 45°-direction almost the double. This is probably due to the preferred crystal orientation and texture inherited in the manufacturing process, i.e. rolling. Furthermore, the 45°-direction shows the lowest mechanical strength of all the tested material orientations. These mechanical observations have to be further investigated and complemented with information of the microstructure, for instance EBSD-technique or FIB.

During the initial phase of the tensile test the material is at first deformed uniform and non-linear with a smooth curvature and homogenously throughout the whole specimen width until 1) is reached in FIGURE 3, [8]. This point indicates the onset of localisation. Thereafter a local defect, edge effect, thinning or large grain is redistributing the deformation in the specimen to a non-uniform deformation i.e. a localisation is initiated. The globally recorded load is thereafter representing the local events ongoing during the propagation of the crack. Furthermore, the continued crack propagation, from point 1) to 3), is linked to the composition of the microstructure and hence guides the path of the crack. During examination of the aluminium foil specimen cross-sections, it can be concluded that almost no plastic deformation is discovered, except for in a small region in the vicinity of the crack plane [3]. Therefore, micro-mechanical understandings of the local plastic effects involved in the fracture processes are needed. This information is necessary to be able to create computer simulation models of the described microscopic events on a macroscopic scale.

A non-linear Hill plasticity model is used in this study to describe the in-plane mechanical properties of aluminium foil. The experimentally measured orthotropic properties of the aluminium foil, according to TABLE 1, are included in the material model to describe the in-plane tensile behaviour. The calibration procedure of the constitutive model is based on the acquired experimental data. Only macroscopic mechanical properties are considered in this work except for the extension of the analytical curve, for a more complete description of how the aluminium foil is modelled at different length scales the reader is referred to [8].

The developed strategy has been verified by means of virtual tension tests. An automated calibration procedure greatly simplifies the work for the CAE engineers working in Abaqus\textsuperscript{TM} [12] as the possible error of assigning an incorrect material model is eliminated and a standardized methodology is obtained.

Plastic strains are implemented with tabular data using the *Plastic card in Abaqus\textsuperscript{TM} [12], with equidistant plastic strain using 0.001 equidistant intervals. The reason why the strains are equidistant is to avoid any automatic regularization that Abaqus\textsuperscript{TM} performs that can result in not appropriate results. The data in the *Plastic card is extended far beyond the experimentally measured failure values using the Ramberg-Osgood adaptation according to FIGURE 4. The main motivation is to allow for localized effects to be modelled with strains larger than the experimental measured macroscopic failure strains. Another motivation for the extension is that Abaqus\textsuperscript{TM} assumes ideal plasticity for plastic strains larger than the specified maximum strain in the *Plastic card and it has been observed by the authors that simulations tend to terminate with an error (failure to converge) for strains larger than specified in this card. The curves have been extended up to several hundred percent plastic straining to enhance the physical description of the material. This assumption has been proved to be valid when moving to micro mechanical simulation models that can capture the realistic cross sections observed in SEM [10].
**Parameterization of the mechanical data**

The experimental test curve can be parameterized as illustrated in **FIGURE 4** with an analytical expression. Furthermore the material model can be analytically calibrated with the Ramberg-Osgood formula, which is given by

$$\varepsilon (\sigma) = \varepsilon^e + \varepsilon^p = \frac{\sigma}{E} + \left(\frac{\sigma}{E_0}\right)^N$$

(1)

Where $E$ is the tensile modulus, which can be obtained directly from the initial slope of the measurements, $E_0$ is the strain hardening modulus and $N$ is the strain hardening exponent. Note that the Ramberg-Osgood formula [1] assumed instantaneous yielding; this implies that the yield surface expands from a point in the origin. The $E_0$ and $N$ parameters can be obtained analytically from the parameterization values. Using the values for strain at break and tensile strength in the Ramberg-Osgood formula we have

$$\varepsilon_{uts} = E \left( \frac{\sigma_{uts}}{E} + \left(\frac{\sigma_{uts}}{E_0}\right)^N \right)$$

(2)

where

$$N = \frac{\sigma_{uts}^2 - 2EW_T}{\sigma_{uts}^2 + 2E(W_T - \sigma_{uts}\varepsilon_{uts})}$$

(3)

and

$$E_0 = \frac{\sigma_{uts}}{\varepsilon_{uts} - \left(\frac{\sigma_{uts}}{E}\right)^N}$$

(4)

$\sigma_{uts}$ is the tensile strength, $\varepsilon_{uts}$ is the tensile strain at break and $W_T$ is the tensile energy absorption. All necessary input to calculate these expressions can be obtained by performing an experimental standard tensile test, thus this is an example of a direct method. The material parameters are visualized in **FIGURE 2**. Furthermore, the in-plane shear modulus $G_{RD\tau D}$ [13] can be calculated with the following expression

$$G_{RD\tau D} = \frac{1}{\varepsilon_{\tau\tau} + 2\varepsilon_{\tau\tau} + \frac{1}{\varepsilon_{RD}}}$$

(5)
Virtual tensile tests

An in-plane material model has been the main focus for the method described here and the out-of-plane properties, not specifically addressed here, are modelled using a transversely isotropic assumption. Measurements of the mechanical properties in the out-of-plane direction are required to calibrate an orthotropic model for solids. The single grain thickness of the foil would most likely provide a stiffer response. Further is the topological thickness variation currently ignored and the out-of-plane Poisson’s ratios are assumed to be identical to in-plane values.

The usage of the material model is verified using virtual uniaxial tension tests in Abaqus/Standard. The method is able to accurately reproduce the physical tests. The verification is conducted to ensure that the materials are correctly implemented and to ensure that boundary effects from the clamping and the effect various parameter reductions in the calibration procedure as described in [1] can be made. Uniaxial tension tests with shell elements are simulated in the same material directions used for the measurements, i.e. RD, TD and 45° direction respectively, as shown in FIGURE 5. The sample dimensions are identical as the experimental test 100 x 15 mm.

The mechanical material behaviour described above is used to describe the average continuum behaviour thus the constitutive numerical material model in the FE-software Abaqus™. Furthermore, the geometrical surface topography has recently been experimentally quantified to more accurately describe the local thickness variation present in the thin aluminium foil, cf. [14]. This local thickness variation is due to the manufacturing principal used. In the last production step is a doubling process utilized. Therefore the aluminium foil has one bright and one matt surface. The bright side is in contact with the work roller and hence the blank appearance and the matt side is due to foil to foil contact during the doubling step. Marks from the working roller surface can clearly be seen on the bright side. Moreover, the matt side has a more stochastic surface topography with higher/deeper peaks and valleys compared to the bright side. The surface topography is regular and smoother on the bright side. This information is used as input for the geometry in the micro mechanical model used to investigate the local effects ongoing during the mechanical deformation in the latter part of the tensile test curve described in FIGURE 3.

Therefore, the mechanical material behaviour and the surface geometry are combined in the micro mechanical models with high spatial resolution. A simulation model result of the cross-section evolution, through the thickness, is presented in FIGURE 6 with the associated constitutive material relationship. The presented approach, with the extended material behaviour has been utilized to capture the realistic local thinning of the aluminium foil in this simulation model. These numerical results can be compared to the experimental observations with SEM made in [10]. However, limited knowledge and information is acquired about the microstructure hence the local information about the grains and grain boundaries. Therefore, this has been identified as a next step to further enhance the details and predictions of the simulation models.

FIGURE 5. Tensile test curve combined with the virtual tensile test curves [8].
Using an analytical expression, in this case Ramberg-Osgood, to parametrize and extending the mechanical data points acquired during experimental measurements i.e. the constitutive material behaviour, is fundamental to create a realistic physical material behaviour for the thin aluminium foil, shown in FIGURE 4. Simulation on the micro meter length scale has made it evident what is related to material constitutive behaviour, microstructure and related to topographical effects. So far, the microstructure i.e. grain structure and grain boundary has not reliably been included in the model. This need more detailed experimental evidence and measurements and simple to use numerical models, currently is Crystal Plasticity, CP, modelling available described by for instance [15]. The mechanical material behaviour and the surface geometry i.e. topography in the cross section, cf. [14], are combined in the micro mechanical model described. The geometry from surface topography measurements and SEM-micrographs of cross sections of aluminium foil was utilized as input for the creation of the geometry of the aluminium foil cross section in the micro mechanical model. SEM images are useful to increase the understanding and to visualize the highly local effects at high magnification and resolution cf. [10]. The mesh density contains several elements through the thickness of the aluminium foil and can readily capture the local effects ongoing through the thickness. The cross section is loaded mechanically in the horizontal direction. A visualization of the cropped simulation model result, zoomed within the same limited area of the cross section, with the corresponding force displacement graph of the cross-section evolution, through the thickness, is presented in FIGURE 7a), b) and c). The deformation sequence in FIGURE 7b) is denoted I) to VII) where I) represents the unloaded initial geometry and VII) is the final geometry when the aluminium foil has failed.

**FIGURE 6.** Micro mechanical model together with constitutive material behaviour with a primary hardening ($H_1$)

**FIGURE 7.** a) Force vs. displacement, b) “cropped” simulation model result with the corresponding local deformation process I) – VII) during mechanical loading of a cross section in aluminium foil and c) summary and overlay of the deformation sequence ranging from I to VII.
These numerical simulation results can be compared to the experimental observations and outputs from SEM made in [10]. The geometry is localized when the force reaches its maximum as shown in FIGURE 7b), between the two different load steps IV and V. However, limited knowledge and information is acquired about the microstructure i.e. the local information about the grains, e.g. size, shape and distribution, and grain boundaries in the aluminium foil. Therefore, this lack of information has been identified as a gap and needs to be further addressed to be able to connect the microstructure with the mechanical performance.

CONCLUSIONS

Experimental results from uniaxial tensile tests have been parameterised with an analytical expression and the slope of the hardening subsequently extended way beyond the experimental data points measured during the experimental tests. This was done in order to accommodate the locally high stresses present in the experiments at the neck formation. An analytical expression, denominated Ramberg-Osgood, is used to describe the non-linear mechanical behaviour. Moreover, it is possible with a direct method to translate the experimental uniaxial tensile test results into useful numerical material model parameters in the general finite element software Abaqus™. In addition to this the extended material behaviour including the plastic flow i.e. hardening, valid after onset of localisation, the described procedure can also capture the true intrinsic material behaviour present at the microscopic events, i.e. geometrical thinning, ongoing in the deformation of the aluminium foil. The engineering sound and parameterised description of the mechanical material behaviour facilitates an efficient categorisation of different aluminium foil alloys and aid the identification of the correct anisotropy (RD/TD/45°) of the mechanical material behaviour based on the physical testing. Micro mechanical simulations have been performed and these numerical simulation results was compared to the experimental observations and outputs from SEM. The cross-section is localized when the force reaches its maximum. However, limited knowledge and information is acquired about the microstructure i.e. the local information about the grains, e.g. size, shape and distribution, and grain boundaries in the aluminium foil. Therefore, this lack of information has been identified as a gap and needs to be further addressed to be able to connect the microstructure with the mechanical performance.

NEXT STEPS

Aluminium-foil could be described with a crystal plasticity framework that has the potential of increasing the accuracy of the predictions with the simulation model results. Constitutive material models and numerical tools including the microstructure of the aluminium foil could be used to link the macro mechanical performance to the current and produced microstructure inherited from the rolling process. This methodology makes it possible to predict the mechanical performance from the material structure. The next step is to include the stochastic grain size and grain shape distribution into the simulation models with the texture and orientation as input.

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1 Introduction
Thin sheets of metals, denominated foil if the thickness is below 200 μm, are frequently utilized in multiple industries such as flexible/stretchable electronics, solar panels and packaging industries to name a few. Aluminium foil (Al-foil) is one of these foils that have been used in a wide range of applications in several industries. It has been found that the mechanical properties of the Al-foil is different to the corresponding properties of Aluminium Alloys that can be found in a general handbook. Several researchers have studied the mechanical properties and the damage processes involved during mechanical loading of different metal foils. For example, [2] have used a simplified constitutive model to predict the onset of localized necking during biaxial stretching of a ductile metal sheet. Later, [3] have used Scanning Electron Microscope (SEM) complemented with a tensile stage and found that small scale stable crack growth occurs and fracture seems to initiate through local necking during deformation in a region that is in the same length scale as the thickness of the Al-foil. It has been found that this behaviour can be described by a modified strip yield model. [4] have studied various examples of Al-foils with thicknesses ranging between

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7-75μm in order to find the influence of the pinholes to the mechanical behaviour. Moreover, [5] continued to measure free-standing sheets of materials consisting of polymer layer and thin Al-foil to further investigate the mechanical continuum material response activated during uniaxial tensile tests in the respective material constituents and laminated sheets of the two materials. The materials and in particular the failed materials/cross-sections were also studied in SEM and the involved micro-mechanisms were investigated. Recently, [6] have extensively investigated thin sheets of aluminium foil with both numerical methods and full field experimental techniques referred as Digital Image Correlation (DIC) or Digital Speckle Photography (DSP). Out of plane deformation (warping) were here concluded as a prominent cause combined with the earlier causes namely strain localization and necking to govern the fracture process.

In the packaging industry recent advancements in the finite element models have called attention to more detailed insights in the micro mechanics and further improved material models for thin aluminium foil, [7–9]. Moreover microstructural information or mapping of microstructure combined with texture and/or local material orientation is needed for the next generations of simulation models to come. Therefore an increased knowledge is needed of the fundamental elastic and non-linear mechanical properties and how the aluminium foil reacts to the prevailing loading scenario during the product life cycle. It has been noticed that, the value of Young’s modulus in a bulk material shows significant different value compared to Al-foil measured by the standard tests method. Young’s modulus of Aluminium Alloys is approximately 70 GPa, as it can be found from the handbook. While several authors, for example [1, 8, 10], got values ranging between 20 and 45 GPa.

Table 1.

<table>
<thead>
<tr>
<th>Temper</th>
<th>Fe</th>
<th>Cu</th>
<th>Zn</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>O</td>
<td>0.7-1.3%</td>
<td>0.05%</td>
</tr>
<tr>
<td></td>
<td>Si</td>
<td>0.05-0.3%</td>
<td>0.05%</td>
</tr>
</tbody>
</table>

### Chemical composition of aluminium AA8079 alloy with a density of ρAl = 2.72g/cm³.

https://matmatch.com/materials/minfc340160-aa-8079

#### 2.2 Twin-Roll production of thin aluminium foil

Manufacturing of Al-foil involves several manufacturing steps. In the final production step, two thin sheets of Al-foil are pressed together and rolled simultaneously (double-rolling), as shown in Fig. 1. The mechanical and surface properties complemented with the associated structure and texture inherited from this manufacturing process may affect the subsequent processes and how the aluminium foil reacts to loads and how the performance will be later on. Two different surfaces, denominated shiny and matt surface respectively, are created during the double-rolling production process. The explanation for this is that the outside surfaces come into contact with the fine polished working roller and therefore become smooth (shiny). On the contrary, the two inside surfaces of the aluminium foils have a slightly more stochastic and roughened surface due to foil to foil material contact and thus these two surfaces appear matt, shown in Figs. 1 and 2. The height of the valleys/peaks on the surface topography is not negligible for the thin thicknesses studied in this work cf. [13]. The manufacturing process generates roll lines from the work rollers on the two external surfaces but are visible on both sides of the foil, although more prominent on the shiny side. A mixture of rolling oil and water is continuously sprayed between the two internal matt surfaces and on top of the two layers, i.e. the shiny sides, during the rolling process to prohibit sticking. Moreover the “twin foil” is continuously rolled down to the desired thickness. After the final rolling step, the foils are separated in a “de-doubling” machine and wounded onto a coil. Most of the Al-foil in this so-called “hard” state is subsequently annealed as a coil in an oven for several
hours. In this process the final grain structure is recrystallized and the microstructure is finalized. Furthermore, the rolling oil is removed in this step, [14].

The final microstructure and texture i.e. process history and manufacturing technique including raw material i.e. chemical composition creates the mechanical characteristics and mechanical performance of the Al-foil described in the next section. Average grain sizes range from approximately 2-10μm for this type of materials studied here depending on the final annealing and alloy used. Therefore, a cross section as shown in Fig. 1, can consist of only one to several grains through the thickness. Furthermore, the microstructure and alloy elements with the precipitates present locally can have a large influence on the local mechanical material behaviour. Due to the thin layer of Al-foil used in the Packaging Industry, ranging from 6μm to 35μm, there is an ongoing discussion if the mechanical properties is similar to the ones found in a bulk material with substantial thicker material or significantly different. One of the aluminium alloys, AA8079, used in the Packaging Industry is described in Table 1.

During making of the foil from a thicker aluminium sheet through repeated rolling, see Fig. 4, the thickness is reduced significantly to a final thickness of 9μm in this study. During the process that requires considerable plastic deformation the volume remain practically constant, which forces the material to expand the covered area in the x1-x2 plane. The plastic deformation occur in a narrow volume in the plate between the two working rollers and more specifically between the two rollers and plate contact areas. The contact areas are stretching across the entire foil in the x2-direction. According to simple geometric relations the chord length and the contact area is around $\sqrt{2hR}$ where $h$ is the thickness reduction during the rolling and $R$ is radius of the roller. This give us the ratio of length to width of the contact area of $\sqrt{2hR}/L$, normally exceeds several orders of magnitude.

The work of sliding friction is proportional to the pressure and length of the contact area. As the roller can be considered to be rigid compared with the softer and yielding aluminium, the plastic deformation requires sliding between roll and the aluminium foil. Since the contact area is stretching the large distance across the foil width, only small segments close to the sheet edges will slide in the $x_2$-direction. Opposed to this the sheet is fairly free to expand in the $x_1$-direction in which the extent of the contact area is small and the frictional forces resisting the slip, becomes comparably small. While the shape of the contact area suggests that the slipping in the $x_2$ direction must be much less than than in the $x_1$ direction the conclusion is that $\varepsilon_{22} \ll \varepsilon_{11}$. The preservation of volume during plastic deformation strain then leads to $\varepsilon_{11} \approx -\varepsilon_{33}$. The constraints should build up a pressure ($-\sigma_{33}$) between the rolls and a pressure ($-\sigma_{22}$) along the contact area while the pressure ($\sigma_{11}$) across the contact area is small. The effect on the mechanical properties suggests that the material becomes isotropic in the $x_2$-$x_3$ plane, while the material properties in the $x_1$-direction may be different due to the preferred alignment of the grains in this direction. Such a material is called transversely isotropic and is a member of the larger group of orthotropic materials.
2.3 Experimental mechanical characterisation of aluminium foil

A severe obstacle is that the tensile test response of thin samples are sensitive to sample preparation and handling, geometrical effects, sample size and imperfections like wrinkles, weak grain, thickness variations etc. Furthermore, sample mounting may sometimes be difficult to accomplish. Tensile tests have been performed in Material Analysis Laboratory in Lund at Tetra Pak® in ten different material orientations, between the Rolling Direction (RD) and the Transverse Direction (TD), since the material behaviour is found to be mechanically anisotropic, [1]. This work focus on the post processing thus analysis of the already available test data acquired from these experimental tests. The definition of material orientation and mechanical characteristics of the thin aluminium foil in RD, TD and Diagonal Direction (DD) oriented 45° to RD, depicted in Figs. 5 and 6. Thereafter, the aquired data, i.e. non-linear mechanical material behaviour, is needed to be further analysed and post processed. This analysis will be further described in the subsequent sections. During this analysis, one specific alloy and type of aluminium foil with a given thickness, 9 µm, has been thoroughly studied, implying that the results may vary for other metal foils. Iron and silicon particles, shown as distributed light gray particles in Fig. 1, are major alloying elements in the 8xxx series which was used in this study. However, the developed procedure and analysis technique may be transferred and used in other industries or to different materials and thicknesses. Furthermore, an extension of basic characterization (RD/TD/45°) were done with an incremental angle of 10° to RD refereed as advanced material characterization, described in [1]. Noticeable in Fig. 6 is that the material directions RD and TD is mechanically deformed to a similar maximum displacement, on the contrary is the maximum displacement in the 45° direction almost the double. This is probably due to the preferred crystal orientation and texture inherited from the manufacturing process, i.e. cold rolling. Furthermore, the 45°-direction shows the lowest mechanical strength of all the tested material orientations.

Experimental measurements were performed on rectangular shaped strip samples with geometrical dimensions 15 x 100mm² in accordance with the ISO Standard 6892-1 (2016) at controlled climate, i.e. 23°C and 50% relative humidity. Totally 45 strips were cut in angles from θ =0°(RD) to 90°(TD) in steps of 10°. In addition to these, 5 strips where cut in 45°(DD) to RD. Fig. 6 shows typical tensile test curves for strip samples for θ =0°, 45° and 90°, that were cut from a coil of aluminium foil. The experimental data, made available online in [1], have shown that the material direction of the foil with relation to the RD affects mechanical behaviour under uniaxial load, shown in Fig. 6. The difference is most profound between the 45° direction in comparison to rolling and transversal direction, RD and TD respectively. At 45° the material has a significant longer plastic region and a smaller yield stress. This is largely influenced by the texture, generally more grains have a higher Schmid factor (soft grains) at 45°. The soft grains would initiate slip and plastic deformation at lower magnitudes of tensile stress making the foil more ductile.

The mathematical analysis of the previously obtained experimental results is an important part needed to be considered to make correct evaluation of experimental test results. An initial study was done by [15] that is extended and enhanced in this work. The aim of this work is to thoroughly characterise the initial stiffness i.e. elastic part of the aluminium foil by the use of the combination of physical testing and mathematical methods. The earlier studies on this specific aluminium foil focused on uniaxial loading in the manufacturing direction i.e. Rolling Direction (RD). In this study, the analysis is extended to the anisotropic properties and material behaviour are further investigated, including DD and TD as well.
3 Mathematical analysis and model based on the experimental data

In this section the analysis of experimental data, analysis procedure, model description and its results are presented in order to characterise the in-plane anisotropic elastic mechanical properties of Al foil. As it is described in the above section about production of Aluminium foil, the material considered here is assumed to be transversely isotropic. Thus, the material properties becomes isotropic in the $x_2 - x_3$ plane, while the material properties in the $x_1 - x_2$ plane, may be different.

As the experiments characterise the properties in $x_1 - x_2$ plane in every $10^\circ$ interval from rolling direction and plus $45^\circ$, therefore tensor properties and its geometrical transformation to the stiffness tensor are used based on the experimental data.

In the following subsections the description of stiffness tensor and its transformation, derived model, and relative results are given.

3.1 The stiffness tensor, compliance tensor and its geometrical transformation

The stiffness tensor is a quantity based on four vectors. Each of these have a geometrical dependence meaning that its components projected on the present coordinate system. Upon rotation of the latter results in dependence of the rotation angle. The basics of the transformation and its consequences are given in this section. The material is assumed to be elastic-plastic strain hardening. The general stress strain relationship writes

$$\sigma_{ij} = C_{ijkl} \varepsilon_{kl},$$

(1)

Here, stress and strain components are denoted as $\sigma_{ij}$ and $\varepsilon_{ij}$ respectively, and the stiffness tensor is $C_{ijkl}$, where $i, j, k$ and $l = 1, 2, 3$.

During the tensile test the stresses are acting in the $x_1 - x_2$ plane which imply that the stress and strain components $\sigma_{13}, \sigma_{12}, \sigma_{23}, \varepsilon_{13}, \varepsilon_{12}$, and $\varepsilon_{23}$ vanish. However, the strain in contraction, $\varepsilon_{33}$, does not vanish. This strain is assumed to be only a few percent and is treated as practically irrelevant during the test. The strain localisation occurring just before failure is not treated in the present investigation. As a result the remaining relevant stress and strain components are $\sigma_{11}, \sigma_{22}, \sigma_{12}, \varepsilon_{11}, \varepsilon_{22}$, and $\varepsilon_{12}$. The strain $\varepsilon_{33}$ is readily calculated from the in-plane stresses once they are known as described in Sec. 2.2.

However, for convenience compliance tensor, which is inverse of stiffness tensor, is used here. The reason why compliance is convenient to use instead of stiffness is explained in the later subsection. Thus, the relationship of stress and strain on incremental form becomes

$$d\varepsilon_{ij} = S'_{ijkl} \sigma_{kl},$$

(2)

where $S'_{ijkl}$ is the compliance tensor.

The coordinates $x'_i$ are directed along the test specimen, in-plane transverse and in the thickness directions. Due to the anisotropy generally the tensile stress will cause shearing. This is manifested as wrinkling of the foil. It is believed that the wrinkling is not impeded by the grips and that the shearing therefore does not introduce any shear stresses. According to the loading conditions the only non-zero stress is $\sigma_{11}$ and the measured strain is $\varepsilon_{11}$. In small regions around the grips the stress state is more complicated. Since the specimens are almost 7 times longer than the width, the deviations from uniaxial tension caused by the grips are ignored in the present study. Thus, the incremental stress and strain relationship is simply

$$d\varepsilon'_{11} = S'_{1111} d\sigma'_{11},$$

(3)

The compliance tensor is expected to have other non-zero compliance components that causes strains $\varepsilon_{22}, \varepsilon_{33}$ and $\varepsilon_{12}$. The specimens are unconstrained with respect to shearing $\varepsilon_{12}$ as already explained. Apart from the small regions close to the grips, also the strains $\varepsilon_{22}$ and $\varepsilon_{33}$ are unconstrained so that the stress field remains uniaxial. Also, slightly asymmetric production conditions are anticipated due to a slight material shear in the $x_2$-direction that increases further from the centre of the roll where the specimen is taken. It is therefore assumed that the principal axes not necessarily have to coincide with the work roller and the rolling direction. A coordinate system $x'_i$ is selected with its axes coinciding with the principal directions of a presumed transversely isotropic material, with the principal directions $x_1'$ and $x_2'$ being identical with respect to material properties, see Fig 7. The non-zero unique components of the compliance tensor are reduced to a canonical set $S'_{1111}, S'_{1122}, S'_{1212}, S'_{2222}$. Symmetry of stress, $\sigma_{ij} = \sigma_{ji}$, and energetic reciprocity lead to additional non-zero components according to $S'_{ijkl} = S'_{ijlk} = S'_{iklj}$.

To obtain the principal compliance tensor components $S'_{ijkl}$ in the principal directions $x'_i$, these are transformed to $x'_i$-coordinates giving the stiffness $S'_{1111}$ for uniaxial stress in
the direction \( x'_1 \) at an angle \( \theta^* \) to the \( x'_1 \) axis, as follows

\[
S'_{1111} = \left( S'_{1111} + 2(S'_{1122} + S'_{1212}) \beta_1 \right) \beta_2^2 + S'_{2222} \beta_4^2 \cos^2 \theta^* - 2(S'_{1122} + S'_{1212}) \cos \theta^* \sin \theta^* + S'_{2222} \sin^2 \theta^*,
\]

where \( \theta^* \) is the specimen orientation and the direction of the uniaxial stress. The variables \( \beta_{ij} \) are the directional cosines defined in two dimensions as follows

\[
\beta_{11} = \beta_{22} = \cos \theta^* \quad \text{and} \quad \beta_{21} = -\beta_{12} = \sin \theta^*.
\]

The cross section strain may be calculated once the in-plane stresses are known, with the following result

\[
de_{33} = S'_{1122} \sigma_{11} + S'_{2222} \sigma_{22} = \left( S'_{1122} \beta_4 \right) \sigma_{11} + S'_{2222} \sin^2 \theta^* \sigma_{22},
\]

where \( \sigma_{11} \) and \( \sigma_{22} \) are the stresses.

The compliance can be calculated as follows

\[
\text{compliance} = \frac{1}{\text{stiffness}} = \frac{d \sigma}{d \epsilon} = \frac{S'_{1111}}{E},
\]

where \( S'_{1111} \) is the stiffness at uniaxial stress and \( E \) is Young’s modulus.

3.2 Mathematical model for calculating compliance at small straining

The compliance is calculated as follows

\[
\left( \frac{d \epsilon}{d \sigma} \right)^{(i+1)/2} = \frac{\epsilon^{(i+1)} - \epsilon^{(i)}}{\sigma^{(i+1)} - \sigma^{(i)}},
\]

where \( i \) is the measurement number, \( \epsilon^{(i)} = u^{(i)} / L_0 \), \( \sigma^{(i)} = f^{(i)} / bt \), \( L_0 \) is the original length, \( b \) is the width and \( t \) is thickness of the sample, and \( u^{(i)} \) and \( f^{(i)} \) are displacement and force respectively. The superscript is the measurement number. The compliance \( (d \epsilon / d \sigma)^{(i+1)/2} \) is taken as the result at the strain

\[
\frac{1}{2} (\epsilon^{(i)} + \epsilon^{(i+1)}).
\]

The compliances, according to Eq (10), for the tensile tests displayed in Fig. 8 are presented in Fig. 9. As observed there is a range of compliances (markers in red) with a linear dependence of the strain. The experimental results are here placed with a linear function of the total strain, \( S'(\epsilon) = \alpha \epsilon + \beta \).

The constants \( \alpha \) and \( \beta \) are fitted to the experimental result as the least square fit with the red marked measurements.
The procedure is here described briefly. The squared differences between \((\frac{d\varepsilon}{d\sigma})^{(i+1)/2}\) and the linear function \(S(\varepsilon)\) are summed, i.e.,

\[
\Omega = \sum_{i=1}^{n-1} \left[ \left( \frac{d\varepsilon}{d\sigma} \right)^{(i+1)/2} - \left( \alpha \varepsilon^{(i+1)/2} + \beta \right) \right]^2,
\]

(11)

and minimised with respect to variation of \(\alpha\) and \(\beta\). Thus,

\[
\frac{\partial \Omega}{\partial \alpha} = 0 \quad \text{and} \quad \frac{\partial \Omega}{\partial \beta} = 0
\]

(12)

This expression lead to the linearly increasing compliance to derive a stress strain relationship. Hence, integration of

\[
\frac{d\varepsilon}{d\sigma} = \alpha \varepsilon + \beta,
\]

(13)

gives the following stress as a function of strain,

\[
\sigma = \frac{1}{\alpha} \ln(\alpha \varepsilon + \beta) + c.
\]

(14)
The integration constant \( c \) is determined by least square fitting to the displacements, as follows

\[
c = \frac{1}{n} \sum_{i=1}^{n-1} \left[ \sigma^{(i)} - \frac{1}{\alpha} \ln(\alpha \varepsilon^{(i)} + \beta) \right],
\]

which readily identifies \( c \) as a simple arithmetic average.

The computed stresses according to Eqn. (14) and its corresponding experimental data for \( \theta = 0^\circ \) are plotted in Figs. 10. The first few measurements presumably represents the complicated mechanics when the mildly wrinkled foil is stretched, in the initial phase of the deformation, that leads to a very low stiffness but decreasing compliance. A perfect test specimen is expected to have only increasing compliance due to increasing plastic deformation. Therefore test results, starting from the first measurement after the inflection point in the tensile test curves, are used. The computed result is used to estimate an as good as possible zero stress and strain point. In Fig. 10 b) the deviation between measured and compiled results for the excluded two markers (blue) before the inflexion point, while the fit between the computed result Eqn. 14 and the experiment (red markers) is almost perfect. The computed strain at zero load is considered to be zero. The strain previous to this is subtracted from the subsequent measurements as observed in Figs. 10 a) and b). Acquired data during experimental tests were taken in equidistant steps of 6μm. The first calculated stiffness is based on stress and strain data from the first two reliable measurements at around 9μm from the estimated unloaded state. For smaller displacements one can only speculate in what happens. However, it seems reasonable to assume that the stiffness is either equal or possibly higher than what is obtained for the region 6μm to 12μm. The first 30 measurements are used to determine the parameters \( \alpha \) and \( \beta \). These are marked red in Figs. 10 a) and b).

3.3 Determination of components of compliance tensor.

In the experimental study by Käck et al. (2015) the focus was on uniaxial test data. Elongation and applied force were registered and are represented in the present investigation as engineering stress and strain unless something else is declared. The meaning of this is that stress and strain are strictly proportional to forces and displacements, scaled per unit of original cross sectional area respectively original specimen length. Displacement refer to the position of the moving grip.

For reasons discussed in Sec. 2.2, the material is assumed to be transversely isotropic meaning that the stiffness quantities \( S_{1111}, (S_{1122} + S_{1212}), S_{2222} \) and the angle to the principal direction \( \theta^* \) that is not known a priori may be obtained from a minimum of four tensile tests, cf. Eqn. (3.1). Tests specimens have to be taken from four different angles to the rolling direction. Four tests are needed because the principal material direction \( \theta^* \) is treated as a fourth unknown.

Numerically the material will never be linearly dependent and as a consequence a solution will always be obtained. However, for a purely isotropic material insertion of (8) and (9) into (6) readily gives

\[
E' = E (\cos^2 \theta^* + \sin^2 \theta^*)^2 = E,
\]

which indicate that for near isotropic materials, the information regarding the principal material direction \( \theta^* \) the quantity \( (S_{1122} + S_{1212}) \) becomes vague, i.e., the equations becomes ill conditioned.

Further, information regarding contraction \( S_{1122} \) or shear \( S_{1212} \) require measurements that involve at least one of them since \( (S_{1122} + S_{1212}) \) is obtained from the tensile tests directly.

The result is 180° periodic and symmetric across the principal directions, which gives the following relations

\[
S_{1111} | \theta' = \theta = S_{1111} | \theta' = 2\theta^* - \theta.
\]
Therefore, the uniaxial compliance or stiffness distribution in the sector $\theta^* - 90^\circ$ to $\theta^* + 90^\circ$ is repeated and reversed across the principal directions and repeated for $360^\circ$.

### 3.4 Anisotropy

To find the elastic stiffness, the inverse of the compliance, is used. In a general case, a matrix inversion is required. However, for uniaxial stress loading the resulting tangent stiffness is simply

$$C'_{1111} = \frac{1}{S'_{1111}}, \quad (18)$$

at any load and especially for vanishing strain the modulus of elasticity is obtained as

$$E'_{1111} = \frac{1}{S'_{1111}} \quad \text{as} \quad \varepsilon_{11} \to 0. \quad (19)$$

Figure 11 shows the variation of the stiffness, Figs. 11 a) and b), obtained at the experiments as functions of total strain. As observed scatter increases with increasing strain. The stiffness calculated from the linearly increasing compliance according to Eqn. (13) is included, see the green curve.

The close-up in Fig. 11 b) covers the region where the stiffnesses are within a 5% accuracy. The red makers are in the region with almost perfect linear compliance versus total strain.

The transformation of the stiffness $C_{ijkl}$, the analogy to Eqn. (3.1), applies to both compliance and stiffness as both are rank 4 tensors. The Eqn. (3.1) is fitted to the experimental result as follows,

$$C'_{1111} = \gamma \cos^4(\theta + \theta^*) + \delta \cos^2(\theta + \theta^*) \sin^2(\theta + \theta^*) + \kappa \sin^4(\theta + \theta^*), \quad (20)$$

where

$$\gamma = C'_{1111}, \quad \delta = 2(C'_{1122} + C'_{1212}) \quad \text{and} \quad \kappa = C'_{2222}. \quad (21)$$

The coefficients $\gamma$, $\delta$, and the angle $\theta^*$ are least square fitted to the experimental result $C'_{1111}$ for different loading angles $\theta$. The small strain limit stiffnesses $E'_{1111}$ are plotted as a function of $\theta$ in Fig. 12. The resulting calculated $E'_{1111}$ are plotted together with the experimental result. The largest stiffness, $C'_{1111}$ is rather close to the rolling direction. The deviation from the rolling direction may be caused by the production conditions during the rolling or post processing of the foil. It is also possible that the placing on the wider original roll that the a few dm wide strip that was used for all test specimens by [1]. Such dependencies was shown for paper cf. [16] e.g.

From the figure one may note that the maximum stiffness 69.6GPa, obtained in the first principal direction at $\theta^* = -16.3^\circ$. Somewhat surprising this is outside the angular range for the tensile tests. The range was selected by [1] as it was assumed a priori, that the principal material direction should coincide with the RD and TD. At the second principal direction at $90^\circ$ from $\theta^*$, i.e. at $\theta^* + 90^\circ = 73.7^\circ$, a second local maximum 63.1GPa is found. In between a minimum stress 58.0GPa is found at $\theta_{\min} = 34.6^\circ$.

The stiffness $C_{2222}$ is in all cases lower than $C_{1111}$, which has the consequence that $\theta_{\min} > \theta^* + 45^\circ$. This follows directly from vectorial properties of the stiffness and the symmetry of $\cos^2$ and $\sin^2$ in Eqn. (3.1) and is confirmed by the experimental result, cf. Fig. 13. Extremum values are expected for principal axes as they also constitute axes of symmetry. The meaning of this is that the solution is mirrored across $-16.3^\circ$ and $73.7^\circ$. It should be noted that the stiffness and the compliance again due to their vectorial properties have a $180^\circ$ periodicity. Fig. 13 summarises principal angles $\theta^*$ to maximum and minimum stiffness.

The anisotropic that is observed for the original material more or less also during the plastic deformation. The transversely isotropic materials’ degree of anisotropy is defined via
Fig. 12. Experimental stiffnesses $E_{1111}'$ at $u_1 = 9 \mu m$ ($\varepsilon_{11} = 9 \times 10^{-5}$) for angles between $0^\circ \leq \theta \leq 90^\circ$ and the least square average ditto plotted in $-20^\circ \leq \theta \leq 90^\circ$. a) all test results b) test results averaged for each measured angle.

Fig. 13. a) the angle $\theta^*$ to maximum stiffness $C_{\text{max}}$, the angle $\theta^*$ to minimum stiffness $C_{\text{min}}$ and the angle $\theta^* + 90^\circ$ to the second local maximum stiffness $C_{\text{max}}$ versus strains from 0 to 4% (cf. Fig. 12).

Fig. 14. Ratio of maximum stiffness to minimum stiffness $C_{1111, \text{max}}/C_{1111, \text{min}}$.

![Graph showing the ratio of maximum stiffness to minimum stiffness.](image)

the ratio of largest compliance versus the smallest. The ratio increases from 1.2% with a peak of 1.42 at the strain 1% and then decreasing to 1.35 at 3% strain (see Fig. 14). It is difficult to define the degree of anisotropy while all specimens have experienced different histories.

This is claimed with the provision that the plastic deformation is path independent. This could possibly affect the results at least at large strains. It should be noted that all compared samples at a specific strain have been exposed to different loads which may cause deviations from the tensor properties after severe plastic deformation.

4 Conclusions

Earlier studies by several authors [1, 8, 10] of thin aluminium foils have given disperse results regarding the elastic properties. The results vary between 20 and 45 GPa and it seems as if the uncertainty emanate from early and almost immediate plastic deformation during mechanical loading.

The observation made in the study is that the foil almost initially for very low mechanical load accumulate plastic strain. By that the compliance is observed to increase linearly to around 1% of straining. The linearity is within deviations of only a fraction of a percent up to the strain 0.5% and then increases to around 5% at 1% of straining.

The material is observed to be anisotropic already from the elastic beginning. Anisotropy quantified as largest compliance versus the smallest is observed to increase around 20% with a peak at around the strain 1% and then falling back about 10%.

The principal directions of the material deviate from the previously a priori supposed rolling direction, cf. [1]. Instead it was found that the principal direction form an angle to the RD in the range from $-16.3^\circ$ for the elastic material and dropping to close to $0^\circ$ at and above 3% strain.
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Micro-mechanisms of a laminated packaging material during fracture

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\textbf{A B S T R A C T}

The micro-mechanisms of fracture in a laminate composed of an aluminium foil and a polymer film are considered in this study. The laminates as well as the individual layers, with and without premade centre-cracks, were tensile tested. Visual inspection of the broken cross-sections shows that failure occurs through localised plasticity. This leads to a decreasing and eventually vanishing cross-section ahead of the crack tip for both the laminate and their single constituent layers. Experimental results are examined and analysed using a slip-line theory to derive the work of failure. An accurate prediction was made for the aluminium foil and for the laminate but not for the freestanding polymer film. The reason seems to be that the polymer material switches to non-localised plastic deformation with significant strain-hardening.

\section{Introduction}

A packaging material commonly consists of several material layers made of paperboard, polymers and aluminium foil. The aluminium foil (Al-foil) and the low-density polyethylene (LDPE) film are studied in this work. Paperboard together with these two materials are widely used as a material structure in aseptic food packages. The Al-foil is used as an efficient barrier towards exposure to oxygen and light in food packages. Furthermore, the Al-foil is usually combined with a ductile polymer layer to extend its durability. Additionally, paperboard layers are added to improve the mechanical strength of the full structure.

The final packaging material is exposed to different loading conditions during its lifetime: forming, folding, filling, distribution, storage, handling and finally opening, wasting and recycling by the consumer. Al-foil is not able to withstand as high local strains as the polymer film and the paper layers. Cracks initiated in the Al-foil can eventually spread into the polymer and the paper layers. Therefore, it is important to understand the individual fracture behaviour of the Al-foil and the LDPE layers and their roles as members of the laminated structure when designing opening devices.

In this work, the focus is solely on the Al-foil and the LDPE film. To be able to predict the damage evolution in the laminate, the fracture behaviour of the Al-foil and the LDPE are at first studied separately. Several studies of the fracture behaviour of the individual packaging material layers, for example paperboard, are presented in [1–4] and a metal film on a polymer substrate used in flexible electronics applications [5–7]. To form a well defined basis for the investigation of Al-foil and LDPE, centre-cracked panels exposed to in-plane uniaxial tensile mode I loading are further analysed in this...
work. Adhesion between the different material layers has not been focused in this study. Therefore, the adhesion has been simplified and idealised. The bonds between the different layers are assigned similar strength as the induced traction forces created when the individual material layers contract due to stress localisation in the tensile tests, thus leading to separation of the two material layers locally. Delamination and the level of adhesion is an intriguing topic, cf. [5–8], and has to be included in future works.

As it was pointed out in [9], mechanical modelling of polymer materials is still in a rather early stage. A cell model has been so far developed and applied to investigate the effect of voids on matrix yielding and localised plastic deformation [10–12]. In the present work, a modified Dugdale model based on the slip-lines that were observed on specimen’s cross-section was applied. This approach was utilised to study the localised plastic deformation of the single layer as well as the laminate. The fracture behaviour of the Al-foil and LDPE laminate has also been studied in previous work [13–15]. A frequent observation is the large variety of involved failure mechanisms in laminates of different compositions. Furthermore, crack tip fields for stationary and propagating cracks have been investigated. The crack tip fields and crack propagation as well as toughening mechanisms in a process zone of a laminate with a stationary crack tip have been investigated in [16].

It has been studied by [17] how the transition to necking can be delayed in polymer-metal laminates. The delay increases the energy-absorbing ability of the structure. The phenomenon is related to the ability of the elastic polymer to maintain a constant tension while the load carrying capability of the metal decreases with increasing deformation. The tendency of the strains to localise in the metal is obstructed by synchronised stretching of the polymer that resists localisation of the deformation. Strain in a periodic laminated structure may localise in regions with a width comparable to the individual layer thicknesses or alternatively in a larger region with a width comparable to the thickness of the entire laminate (cf. [18]). The study also suggests suitable combination of materials and layer thicknesses that will improve the structural toughness. The analysis suggests that a band of interacting co-necking layers for a tilted band across the laminate can arise during necking of several interacting individual layers.

1.1. Motivation, focus and aim

Opening devices have in recent years significantly increased in volume in the packaging industry. The failure process during the opening is intended, therefore the damage initiation and propagation of the process leading to complete fracture has

<table>
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<th>Nomenclature</th>
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<tr>
<td>$a$</td>
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<tr>
<td>$h_{crA}, h_{crL}$</td>
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<td>$m_A, m_L$</td>
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to be controlled, cf. [19]. The obtained results from this study will be used in the development of new packaging materials including customised opening devices.

Micro-mechanical understanding of the local plastic effects involved in the fracture processes are in focus in this work. This information is necessary to be able to create accurate and reliable computer simulation models of the described microscopic events on a macroscopic scale. Selection of appropriate constitutive models for the continuum material and how the crack initiates and propagates to various loading conditions can also be simplified. Hardware and software improvements have emerged during the last decades. Therefore the demand has significantly increased during the last years to provide efficient and sufficiently simple tools to solve industrial applications where failure initiation and subsequent crack growth need to be accurately predicted.

The aim of this work is to present results of the mechanical and fracture mechanical response of the two studied materials, Al-foil and LDPE. This knowledge will be transferred and used as input to virtual material models for prediction of the different package functionalities. The aim is that the presented approaches should be utilised to create an efficient and powerful tool for decision support and to be able to drive the package development with realistic and predictive simulation models.

This paper consists of three parts: the first, an experimental part, describing the studied materials and the performed experimental tests. The second part, visualisation of the involved micro-mechanisms, is introducing the sample preparation technique used for creation of micrographs and finally the third, theoretical part. In Section 2 the material properties are given. Experimental setup and experimental results are also presented in this section. Section 3 describe the capturing of scanning electron microscope (SEM) images and the post test observations. Section 4 describes the equations of the micro-mechanical processes using a slip-line theory and a micro-mechanical approach are used to compute the mechanical behaviour of the macroscopic crack. The analytical models are used to examine the experimental data.

2. Mechanical tests

2.1. Materials

A 9.0 μm thick, fully annealed AA1200 Al-foil and a 27.0 μm thick LDPE polymer film with the product name LD270 are studied in this work.

To simplify the analyses both materials are treated as homogeneous and isotropic elastic–plastic. The following properties have been measured for the studied Al-foil, modulus of elasticity $E_A = 55.6$ GPa and ultimate tensile stress $\sigma_{uA} = 73$ MPa. The yield stress is not easily determined. The lowest possible stress at which plastic deformation occurs is estimated to be at $\sigma_{YA} = 36$ MPa [20]. The elastic modulus is low compared to what is expected for the bulk material of aluminium. A large variety of studies using direct tensile testing of thin Al-foils, optical methods for measuring displacements, indentation tests etc. obtain a large variation of moduli ranging from 25 GPa to slightly above 70 GPa [21,22]. In spite of the variety of experimental results, mostly showing much lower elastic moduli than the bulk value, a physical motivation for the deviation remains to be found. In the present study the bulk value of the elastic modulus for aluminium is chosen to $E_A = 71$ GPa.

For the LDPE, the modulus of elasticity is measured to be $E_L = 126$ MPa with the ultimate tensile stress $\sigma_{UL} = 8$ MPa. The yield stress is estimated to $\sigma_{YL} = 5.3$ MPa which is in accordance with the result presented in [23]. Poisson’s ratio is not easily measured. Commonly used are values around 0.3 as for many metals. A $\nu_L = 0.33$ is used for Al-foil (cf. [13]) and LDPE is known to be close to incompressible with very high Poisson’s ratios, e.g. in [24] $\nu_L = 0.45$. As opposed to this [25] use $\nu_{UL} = 0.3$ for a laminate where the contraction is dominated by the LDPE. Since the laminate is composed of around 10% Al-foil and the rest is LDPE the conclusion is that Poisson’s ratio is selected to $\nu_L = 0.3$ for the LDPE. The Young’s modulus of the laminate, $E_{lam} = 17.9$ GPa, can be accurately estimated by the theory described in [26]

$$E_{lam} = \frac{(m_L + m_A)^2 - (v_L m_L + v_A m_A)^2}{(h_L + h_A)(m_L + m_A)},$$

where

$$m_L = \frac{E_L h_L}{1 - \nu_L^2}, \quad m_A = \frac{E_A h_A}{1 - \nu_A^2},$$

(2)

where $h_A$ and $h_L$ are the initial thicknesses of Al-foil and LDPE respectively.

2.2. Tensile tests

Thin centre-cracked sheets, as shown in Fig. 1, were used for evaluating the fracture mechanical performance of the studied materials. The polymer-coated sheet and the individual material layers of Al-foil and LDPE were experimentally tested according to the ASTM (D-882-91) convention [27].

Pre-fabricated cracks were manually cut using a sharp scalpel. Crack lengths ranged from $2a = 2$ mm to 45 mm. The width and gauge length of the specimens was $2W = 95$ mm respectively $2H = 230$ mm. The testing was made at room temperature in a MTS Universal Testing Machine in the Laboratory facilities at Blekinge Institute of Technology. The upper clamp
is attached to a 2.5 kN load cell. After positioning the specimen, the upper and lower clamps were closed. Pressure was applied to tighten four equally spaced quick-acting locking nuts along the front of each clamp. The specimens were extended by traversing the upper cross-head at a constant speed of 7 mm/min. During testing, the position of the cross-heads and the applied load were monitored and recorded. The tests of the specimens were run until the entire cross-section was fractured. Experimental results from the individual layers of Al-foil and LDPE are shown in Fig. 2 for a selection of crack lengths.

Average results from five tests of each case are presented in the two graphs. Engineering/nominal quantities are displayed on the $x$- and $y$-axis. The tensile strength systematically decreases with increasing crack lengths. Implementing a centre-crack in the specimen enables repeatable results from the fracture mechanical testing.

The experimental results from the one-side laminated Al-foil with LDPE are showed in Fig. 3(a). All curves show a considerable difference in the slope at the point where the Al-foil layer is completely broken. Circular markers show the position where this occurs in the two graphs, denoted with IV. In the laminate, there is no sudden drop in the load curve after the Al-foil layer is completely failed. This is due to the locally stretched LDPE layer which is able to withstand such high loads at the local high strains. A considerable strain hardening is occurring in the polymer. If the adhesion is weak or the two materials detaches and start separating a flat and horizontal part is observed in the curve. In most of our studied samples the slope was changed when the Al-foil has failed completely, indicating a sufficient adhesion level. The direction or inclination of the slope was defined by the amount/width of LDPE in the cross section and also the strain hardening in the localised neck that is formed in the area where Al-foil has failed.

![Fig. 1. Experimental test specimen and definition of symbols, to the right is a completely broken specimen exerted to the tensile tests.](image1)

![Fig. 2. Stress vs. strain from experimental uniaxial mode I tensile tests for a selection of crack lengths (a) Al-foil [20] and (b) LDPE [23].](image2)
A summary of the experimental results for centre-cracked specimens with an example of the crack length $2a = 45$ mm is depicted in Fig. 3(b). In Fig. 3(b) a sheet of Al-foil and LDPE have been put together with adhesion (Al-foil + LDPE) and without adhesion (Al-foil/LDPE) included. The latter resulted in the curve “independent layers” as shown in Fig. 3(b). Fig. 4 is an illustration to visualise the realistic deformation occurring close to the vicinity of the crack tip during the experimental tensile test. All three variants are displayed, hence both of the two individual material layers and the laminate. Five different stages (I, II, III, IV, V) during the deformation of the laminate is indicated in the load curve in Fig. 3(b).

The differences between the load carrying capacity of the laminate and that of the individual material layers are surprisingly large as described in [13]. The raw material, the manufacturing technique and the manufacturing process utilised at the production site introduce and create the micro-structure in the thin Al-foil. The grain size, grain borders and surface topology influence the mechanical response to a large extent. When the Al-foil is laminated with a ductile polymer some of these effects are decreased. It was suggested by [13] that surface irregularities that are observed in the untouched Al-foil is covered by the LDPE layer. It was suggested that this could contribute to the increased strength of the test specimens without a crack. However, the increased strength of the cracked specimens as observed here seems less important. Surface cracks and defects introduced during manufacturing of the Al-foil due to its thinness arrest in the Al-foil when laminated with a LDPE layer. Therefore, the mechanical behaviour in the laminated Al-foil is less sensitive compared to the freestanding counterpart. Hence the two materials co-evolve in the laminate and they are observed to increase the load carrying capacity together. Furthermore, LDPE delocalise and redistribute the load in the local thinning of the Al-foil and prevents surface cracks and defects in the Al-foil to dominate the mechanical behaviour.

Fig. 3. Stress vs. strain from experimental uniaxial mode I tensile tests for a selection of crack lengths (a) One-side laminated Al-foil and (b) load vs. extension summary of the Al-foil, the LDPE and the laminate with a centre-crack $2a = 45$ mm [13].

Fig. 4. Crack tip opening in the vicinity of one of the crack tips during the progressive failure for the Al-foil, LDPE and laminate layers. Five stages of the deformation in the laminate layer is shown (I) crack initiation (II) during crack propagation (III) Al-foil approximately broken half way towards the free edge (IV) Al-foil is completely broken in the laminate (V) LDPE approximately broken half way through the material in the laminate. The notation I–V is explained in Fig. 3(a) and (b).
Fig. 3(b) is presenting the experimental results with the load vs. extension instead of stress vs. strain. A stiff and strong material such as Al-foil is laminated with a soft and weak polymer film, in this case consisting of LDPE. The load carrying capacity, as compared with that of the individual layer of the Al-foil, has almost doubled for the laminate. This is a bit surprising since the freestanding LDPE at the same extension carries around a tenth of the maximum load of the laminate. The materials co-evolve to deliver maximum load in the laminate at a displacement where both as individual layers would carry almost no load. At 2% strain the LDPE is expected to carry only 2 or 3 N and the Al-foil would be broken.

3. SEM investigation

3.1. Tools and procedures

Microscope images were created to get a better understanding of the involved micro-mechanical processes during the failure of the packaging material. The procedure of creating SEM images was obtained in three steps. In the first step a mikrotome cutter from the manufacturer Leica was used. The machine cuts 50 µm thin slices along the direction AA to BB depicted in Fig. 1. The cutting blade that was used is believed and has by experience shown to be sufficiently sharp so that undesired influence is avoided. Furthermore, two of the specimens are tilted during cutting to obtain a three dimensional effect in the SEM micrograph, cf. Fig. 5. The second step was to cover the specimen with a coating of a few nm of gold particles. This was done in a Cressington 108 auto sputter coater. The final step was to analyse the prepared test specimen in the SEM, and to create pictures with a fine resolution. The SEM equipment is a Hitachi-Tabletop Microscope, TM-1000 operating at 15 keV. This experimental setup is available in the Laboratory at Tetra Pak® Packaging Solutions AB in Lund.

3.2. Post-test examination of specimens

The LDPE and the Al-foil material layers display similar response curves with respect to the overall mechanical behaviour when a centre-crack is introduced in the test specimens, presented in Fig. 2(a) and (b). However, the deformation mechanisms in the two materials occur at different stress and strain levels. This behaviour includes thinning of the cross-section accompanied material softening when the crack propagates. Furthermore, the fracture process in the Al-foil is related to how the crystals/grains are built up and in the LDPE there is a combination of how the molecular chains are arranged and how the structural arrangement of the amorphous zones, crystallographic slip-planes and crystallites are mixed and organised [28,29].

In addition to this, at the examination of the Al-foil specimen cross-section, almost no plastic deformation is discovered, except for in a small region in the vicinity of the crack plane. The view in Fig. 5 is the upper half of a cut through the specimen as is marked AA – BB in Fig. 1(b). The plastically deformed region is confined to a region with a linear extent of 20–30 µm. The region is marked as a dashed arrow head in Fig. 5(b). The arrow tip represents the crack in the plane of the cut (see AA – BB in Fig. 1(b)). Also the LDPE shows localised plastic deformation in a small region close to the crack surface in the vicinity of the crack tip region. In Fig. 5 it is observed that the thickness of the LDPE very little reduced outside the region marked with dashed lines. Examination of the edge of the fractured LDPE and Al-foil, using SEM, provides visual evidence that failure occurs through plastic localisation.

The post-fracture examination of the Al-foil specimen cross-section shows almost no plastic deformation except for in a small region in the vicinity of the crack plane, as shown in Fig. 5. One of the fractured test specimen half is shown in the two figures in Fig. 5 and the section cut is through the thickness direction parallel to the loading direction (AA – BB in Fig. 1).

However, the maximum load carrying capacity and the flexibility are very different in the two materials. Nevertheless the two materials act similarly in the cross-sections close to the crack tip. Examination of the edge of the broken LDPE and

![Fig. 5](image-url) Micrographs of fractured cross-sections stretched in the y-direction (mode I) of (a) single LDPE and (b) Al-foil layers respectively. The sample has been tilted in the SEM equipment to enhance the 3D.
Al-foil, using SEM, provides visual evidence to the assumption that process is governed by localisation and plastic deformation. This localisation occurs in a very small strip region of the test specimen during the tensile deformation. The cross-sectional area in this necked region decreases by a larger proportion than the locally stretched material manage to strain hardens [30,31].

Fig. 6 shows a laminate during deformation process consisting of a double-side LDPE coated Al-foil. For this case the matching parts of the fractured Al-foil could be found while an unbroken coating on one side held the specimen together. The fractured cross-sections are similar for the case of the Al-foil compared to a single-sided polymer coated Al-foil Fig. 7 and the freestanding Al-foil, presented Fig. 5(b). The fractured Al-foil is observed to interact with the surrounding polymer. It is evident that the Al-foil and the LDPE layer both fail through localised plastic deformation. Localised plastic deformation is also observed to spread into the unbroken LDPE layer.

A single layer of the LDPE film withstands a rather long extension during loading. When laminated with the stiff and strong Al-foil, LDPE is enforced to localised deformation in the region where Al-foil fractures. LDPE is exerted to high strains and stresses locally in this created band. Furthermore, this newly created band is only a few microns wide in the beginning. This phenomenon occurs if the adhesion level is sufficiently high and hence delamination between the different material layers is suppressed. This effect increases the load carrying capacity due to orientation and strain hardening in the locally stretched LDPE. Therefore, the polymer substrate carries load that is many times the load expected at the nominal straining at fracture (see Fig. 3). The adhesion between the two layers also enforces the dissipation to cover both the fracture of the Al-foil and the LDPE. In-situ tensile testing complemented with microscopic pictures of the cross sections is needed for increased understanding of the involved micro-mechanisms in the sequence of deformation during the fracture process.

4. Theoretical model

4.1. Local plasticity model

The theoretical model that is used to compute the fracture process is based on an elastic–plastic von Mises material model. However, it is possible that the semi-crystalline polymer LDPE has a more complicated behaviour compared to Al-foil, e.g., due to a locking effect that longer molecular chains might cause during excessive straining [9–11]. Furthermore, considering the tensile test result observed in Fig. 2, that shows a roughly similar behaviour for both LDPE and Al-foil in the strain region of interest, the same model is used for both materials. The maximum stress occurs at 0.04–1.3% nominal strain in the Al-foil and from 10% to 20% nominal strain in the LDPE which is considered to be small to fairly small strains (cf. Fig. 2).

According to the von Mises elastic–plastic material model, the plastic deformation that is initiated in a uniaxially stretched thin film/foil immediately localises to a narrow band across the material at the thinnest cross section. At uniaxial tension, the band forms a straight line that is at a material dependent angle to the pulling direction [32]. The von Mises yield criterion is fulfilled in the band and plastic deformation occurs according to a flow rule. The width of the band is initially close to the original thickness of the foil. During stretching, the band widens and the local thickness is rapidly reduced. When the width of the band has reached twice of its initial width, the thickness is reduced to zero. Simultaneously the load carrying capacity reduces and vanishes when the cross-section disappears.

In the neighbourhood of a crack tip, a consequence of the flow rule is that the strains at first are more or less homogeneously distributed. At this stage the stress field ahead of the crack tip is mostly biaxial which prevents the strain from localising. During increased remote load the crack gets blunted [33] and as this widens the crack tip, the stress parallel with the crack plane is gradually reduced. When this stress is half of the stress perpendicular to the crack plane, the condition for strain localisation is fulfilled, cf. [32]. After this the strains localise to a band continuing ahead from the crack tip as observed.

![Fig. 6. Micrographs of localised plastic deformation in the Al-foil (mode I) of double-sided polymer coated Al-foil. (a) Initiation of the localisation and (b) complete failure of the Al-foil layer.](image-url)
by Dugdale [34] who artificially manufactured a blunted crack tip. He observed that the best result was obtained by using cracks with the same width as the thickness of the specimens.

Fig. 8 shows cross-sections with developing bands of localised straining. The structure is a laminate consisting of a stiff layer, Al-foil, bonded to a weak layer, LDPE. The crack is situated in the $x-z$ plane (see Fig. 1). In the band the plastic deformation occurs as a slip-line along planes that form a $45^\circ$ angle to the specimen surfaces in the $y-z$ plane. A continuous change of the position of the slip-planes leads to the geometries shown in Fig. 8. The adhesion between the two layers is assumed to be strong, hence the interface remains intact in the homogeneously deformed parts of the Al-foil, but not strong enough to prevail in the region between $A$ and $B$ in Fig. 8(a), where the strains localise. The stiff layer is supposed to deform more or less independently of the behaviour of the weaker layer. The localised plastic deformation in the stiff layer introduces large strains in the weak layer that forces it to large deformation locally. Because of these assumptions, the width of the band of localised straining in the weak layer along the interface is the same as the thickness of the stiff layer [35].

The laminate is subjected to pure tension in the $y$-direction while the stresses in the $z$-direction are expected to be insignificant. Further the variation in the $x$-direction is assumed to be small over distances that are of the order of the laminate thickness. Therefore, plane strain is expected in the $y-z$ plane. Since the plastic strains dominate, the plane strain condition applies directly to the plastic straining, that is inhibited along the band of localised strain. This creates a tensile stress in the $x$-direction that approaches $\sigma_y/2$, according to the flow rule of von Mises yield condition (cf. [36]). For a single layer, the von Mises effective stress, $\sigma_e$, and the stress at break, $\sigma_b$, becomes

Fig. 8. Specimen cross-section and slip-lines of the two layers of the laminate with a broken interface in the region of localised plastic deformation; (a)–(d) show the separated layers defining the reduction of their cross-sections during deformation and damage process. The materials are stretched in $y$-direction.
\[ \sigma_b = \sigma_c = \sqrt{\sigma_y^2 + \sigma_z^2 - \sigma_y \sigma_z} = \frac{\sqrt{3}}{2} \sigma_y \approx 0.866 \sigma_y, \]  
and thus, the force, \( F \), per unit of length becomes
\[ F = \sigma_y t = \frac{2 \sqrt{3}}{3 \sigma_\delta t} \approx 1.15 \sigma_\delta t, \]  
where \( t \) is the actual thickness.

For a laminate, the original thicknesses are \( h_A, h_L \) respectively, the corresponding force per unit of length carried by the actual thickness in the localised strain region in a laminate becomes
\[ F = \frac{2 \sqrt{3}}{3} (\sigma_{\delta A} t_A + \sigma_{\delta L} t_L), \]  
where \( \sigma_{\delta A} \) and \( \sigma_{\delta L} \) are the stresses at break of the respective Al-foil and LDPE layers, \( t_A \) and \( t_L \) are the actual Al-foil and LDPE thicknesses in the region of localised strain.

Furthermore, consider an Al-foil with the original thickness \( h_A \) and actual thickness \( t_A \) (see Fig. 8(b)). The volume per unit length of the material is supposed to be constant during the plastic deformation. Initially the volume of the plastically deforming segment is \( V_0 = h_A^0 \). The volume is readily given by the geometry in Fig. 8(b) as \( V = (h_A + \delta - t_A)(h_A + t_A)/2 + t_A^2 - h_A^2 = \frac{1}{2} (\delta + t_A - h_A)(t_A + h_A) \). The change of volume \( V - V_0 \) is as follows,
\[ V - V_0 = (h_A + \delta - t_A)(h_A + t_A)/2 + t_A^2 - h_A^2 = \frac{1}{2} (\delta + t_A - h_A)(t_A + h_A). \]  
Thus, \( V = V_0 \) only have one permissible solution for \( t_A = t_A(\delta) \), which according to (6) is,
\[ t_A(\delta) = h_A - \delta. \]  
The slope of the deformed segments is given by the geometric relations in Fig. 8(b). Along the surface segments one obtains,
\[ \frac{dz}{dy} = \pm \frac{h_A - t_A}{h_A + \delta - t_A} = \pm \frac{1}{2}. \]  
The deformation of the LDPE layer is modified with regard to the initial slip-lines. Because of the influence of the stiffer and harder Al-foil the initial position of the slip-lines is the position shown in Fig. 8(a). The assumption is that the interface between the laminate layers breaks in the band of localised straining (see Fig. 6). Therefore a simultaneous two-sided thinning of both the Al-foil and the LDPE layer occurs. To maintain continuity across the interface of the extension both layers will give identical width of the localised plastic zone on both sides of the interface (see Fig. 8). When the Al-foil breaks the slip-lines meet on the upper part of the LDPE as in Fig. 8(c) point C. After this the rate of motion of the point where the slip-lines meet is doubled giving a slope of the surface segment \( dz/dy = \pm 1 \). To maintain constant volume rate along the simultaneously created surface segment on the lower part of the LDPE, the inclination of the created surface should be \( dz/dy = \pm 1/3 \), which is obtained by using geometrical considerations. The moment the LDPE and thereby the entire structure fails is displayed in Fig. 8(d).

The surface profiles shown in Figs. 5–7 give confidence to the assumption that the fracture occurs through localised plastic straining. In Fig. 5 the slope 1:2 that is included in the SEM image shows a fair fit to the image of the fractured Al-foil. The correspondingly included theoretical profile for the LDPE is also reasonably similar to the experimental result. One has to consider that the LDPE is a soft and flexible material and the fractured cross-section shape may have been slightly deformed and folded during the specimen cutting, preparation and mounting which may partly account for the discrepancies between the theoretical and the experimental results. Polyethylene was reported in [37] to be dominated by shear strain deformation. Local plastic deformation leading to complete separation of a metallic foil of Al-foil was observed by Dugdale [34] and the same for other metallic materials by several other investigators after him [37]. The essence of the observation is that the fracture process of thin foils of many different materials is essentially a purely plastic process, that depends on the elastic–plastic material properties and the layer thickness. The lack of stress constraint because of insignificant stress across the plane of the foil causes the foil to deform plastically at a stress on the level of the yield stress. That prevents the development of the high stress that is required for cleavage fracture or initiation of voids for ductile fracture. The plastic deformation leads to purely plastic failure before the work of failure of the material is reached. This means that the macroscopically observed critical stress intensity factor is, as opposed to the work of failure, not a material property.

As observed in Fig. 5 the cross-section is thinned to vanishing thickness as an effect of the plastic deformation for both the Al-foil and the LDPE. Under these circumstances, it seems reasonable to assume that the final breaking of the cross-section has very little effect on the process as a whole. The suggested type of the fracture process based on pure plastic deformation depending on a strong coupling between the individual layers of laminates, has been suggested by [13].

For both layers the geometrical relation between the layer thickness and the extension \( \delta \) of the region becomes
\[ t_A(\delta) = h_A - \delta \quad \text{and} \quad t_L(\delta) = h_L - \delta \quad \text{for} \quad \delta < h_A, \]  
\[ t_A(\delta) = 0 \quad \text{and} \quad t_L(\delta) = h_L - \delta \quad \text{for} \quad h_A \leq \delta < h_L. \]
and
\[
t_A(\delta) = t_L(\delta) = 0 \quad \text{for} \quad h_L \leq \delta.
\]

The force per unit of length along the $x$-direction according to (5) becomes
\[
\begin{align*}
F &= \frac{1}{\sqrt{3}} \sigma_{ba} h_A + \sigma_{bl} h_L - \theta (\sigma_{ba} + \sigma_{bl}) \delta \quad \text{for} \quad \delta < h_A, \\
F &= \frac{1}{\sqrt{3}} \sigma_{bl} (h_L - \delta) \quad \text{for} \quad h_A \leq \delta < h_L, \\
F &= 0 \quad \text{for} \quad h_L \leq \delta.
\end{align*}
\]

The relation to the spatial coordinate ahead of the crack tip requires a solution of linearly elastic problem for the body. Such a solution can be achieved analytically using an integral equation method as in [38] or by application of a numerical finite element method. The cohesive stress is then applied as boundary conditions of a Robin type. Fig. 9 shows the cohesive force per unit length as a function of the displacement according to (12). The kink connecting the two linear segments indicates when the thinner Al-foil is discontinued.

4.2. Fracture mechanics analysis

Consider a laminated sheet consisting of one layer of Al-foil and one layer of LDPE containing a premade centre-crack. A coordinate system is attached to one of the crack tips as shown in Fig. 1. The crack tip, give the coordinates of the sheet as $|x + a| < W$, $|y| < H$ and $|z| < h/2$, where $h = h_A + h_L$. The initial position of the crack is $|x + a| < a$, $y = 0$ and $|z| < h/2$. A cohesive stress in unit length $\sigma_0(\delta) = F(\delta)/h$, where $\delta$ is the discontinuous displacement between the upper and lower surfaces of the region $0 < x < \rho$, $y = 0$ and $|z| < h/2$, cf. Fig. 10(a). The cohesive force, $F(\delta)$, is given according to Fig. 9. The developing region of localised strain is observed in the $y-z$ plane in Fig. 10(b) for the individual layer of Al-foil and in Fig. 10(c) the corresponding deformation for the laminate is showed.

By considering that the in-plane strains $\epsilon_x$ and $\epsilon_y$ are equal in both layers one may replace the individual elastic parameters with an equivalent modulus $E_{\text{lam}}$ and an equivalent Poisson’s ratio, $\nu_{\text{lam}}$, that then become applicable to the laminated structure as if it was a homogeneous material. The equivalent cohesive force per unit length is as it is given by (12).

The mathematical solution for arbitrary cohesive forces may be solved using Muskheilishvili’s method [39]. The obtained integral equation may be solved using an iterative method because of the mixed boundary condition in the cohesive zone, where $F = F(\delta)$, cf. [38]. At small scale yielding, i.e. when $\rho$ is much shorter than the length of the crack and the distance to the traction free boundary, the critical value of the $J$-integral [40] is directly given by the work required to break the Al-foil [41]. The following relation, readily computed from evolving geometry of the cross-section, gives the critical value of the $J$-integral, $J_f$, as, cf. [42],
\[
J_f = \frac{1}{h_A + h_L} \int_0^{h_A + h_L} F(\delta) d\delta = \frac{1}{\sqrt{3}} \left( \frac{\sigma_{ba} h_A^2 + \sigma_{bl} h_L^2}{h_A + h_L} \right).
\]

Table 1 shows the calculated values for the Al-foil, the LDPE and the laminate consisting of a layer of Al-foil and one layer of LDPE. The work of failure $J_f$, as calculated from (13). Also the critical $J_f$ of the individual layers are computed using (13) by putting the thickness of the respective counterpart to zero.

![Fig. 9](image) Force in the $y$-direction per unit of length in the $x$-direction versus displacement across the band of localised strain. The force represents the load carrying capacity of the band of localised strain.
To compute the small scale yielding result for the limiting stress depending on the crack length the result by [43] is used.

The following gives the critical stress $\sigma_c$ at initiation of crack growth

$$\sigma_c = \sqrt{J_f \frac{E}{\pi a} \phi \left( \frac{a}{W} \right)},$$

(14)

where

$$\phi \left( \frac{a}{W} \right) = \frac{1 - 0.025 \left( \frac{a}{W} \right)^2 + 0.06 \left( \frac{a}{W} \right)^4}{\sqrt{\cos \left( \frac{\pi a}{W} \right)}},$$

(15)

cf. [43]. The critical stress for different centre-cracked specimens has been multiplied with the cross-sectional area in order to predict the critical force in Fig. 11. A comparison of the maximum force from the experimental data versus the derived

**Table 1**

Comparison of structural and material parameters for the different test specimens.

<table>
<thead>
<tr>
<th></th>
<th>Al-foil</th>
<th>LDPE</th>
<th>laminate</th>
</tr>
</thead>
<tbody>
<tr>
<td>$h$ (µm)</td>
<td>9.0</td>
<td>27</td>
<td>36</td>
</tr>
<tr>
<td>$E$ (GPa)</td>
<td>71.0</td>
<td>0.126</td>
<td>17.9</td>
</tr>
<tr>
<td>$v (-)$</td>
<td>0.33</td>
<td>0.45</td>
<td>0.30</td>
</tr>
<tr>
<td>$\sigma_b$ (MPa)</td>
<td>73.0</td>
<td>8.0</td>
<td>26.6</td>
</tr>
<tr>
<td>$J_f$ (N/m)</td>
<td>188</td>
<td>82.6</td>
<td>109</td>
</tr>
<tr>
<td>$F_{max}$ (N) ($2a = 45$ mm)</td>
<td>14.0</td>
<td>9.4</td>
<td>24.4</td>
</tr>
</tbody>
</table>

**Fig. 10.** Strip yield zone ahead a crack tip. (a) The crack geometry in the plane $z = 0$. (b) The slip-region as seen in a plane $x = \text{const.}$ in the region $0 \leq x \leq d$ for the Al-foil and (c) in the laminate.

**Fig. 11.** Force vs. crack length of LDPE, Al-foil and Laminate [44].
expression presented above is shown in Fig. 11. Eq. (14) is accurate both for freestanding Al-foil and a laminate consisting of one-side laminated Al-foil with LDPE. The equation is used for the freestanding layers by inserting zero thickness of the absent layer. Furthermore, the analytical expression calculating the critical force is not predicting the behaviour of LDPE-film satisfactory. The result is an underestimation, in this case, as indicated by the dashed line in Fig. 11.

5. Results and discussion

In a laminate, the fracture process is localised in the LDPE and the material is exerted to plastic deformation and high stresses close to the crack tip. The reason seems to be that the stiffer Al-foil enforces localised plastic deformation in the polymer layer at locally high strains probably due to strain-hardening and re-orientation of the molecular chains. This results in increased load carried by the LDPE around four to five times larger in the laminate than it is at the same extension as a freestanding film, cf. Fig. 3(b). This explains why LDPE solely can withstand such a high loading/force, the “tails” in Fig. 3(a). Al-foil has fractured completely in the region where the slope of the latter part of the curve change direction significantly. Therefore only the LDPE layer is locally carrying the high loads in the tails of the graphs. The adhesion and the interaction, especially the “load sharing” i.e. delocalisation of stress in Al-foil and localisation of stress in LDPE, between the packaging material layers significantly increase the peak load in a laminate.

While a freestanding polymer film made of LDPE can be extended to a high degree before breaking, the fracture in the polymer film is not localised due to the low stresses that is emerged in this case. Al-foil laminated with the same LDPE-film shift the fracture mechanical behaviour of the LDPE layer completely. LDPE is in such a case forced to fail simultaneously with the Al-foil if the adhesion level is sufficient to prevent delamination from spreading along the interface between the material layers. Hence the stress in the LDPE layer is localised in the region where Al-foil has ruptured.

Plane stress conditions and specimens with thicknesses larger than a few millimetres requires a comparably large work of failure, often larger than the work of fracture at plane strain conditions. For very thin specimens like in the present case the failure mode switches from fracture to localised plastic deformation in a region that is of the order of a few microns. It is observed from Eq. (13) that the failure mode switches to fracture when the thickness exceed a critical value. Assuming a fracture toughness of 24 MPa \( \sqrt{m} \), which is reasonable for a bulk aluminium material, the critical thickness according to (13) becomes \( h_{\text{cr}} \approx 190 \mu m \) or 21 times the studied thickness. In the analysis it is the “far-field” \( J \)-integral that is calculated since the plastic deformation outside the slip-region is assumed to be very small and negligible. This is in accordance with the observations made during the experiments for Al-foil and the laminate but not for the stand alone LDPE layer which could explain why the model fails in the latter case.

A slip-line model is presented for the localised plastic deformation of the freestanding and laminated Al-foil. The model can successfully be used to predict the final shape of the failed cross section of both the freestanding and laminated Al-foil, cf. Fig. 11. Fig. 5 shows the local deformation in the vicinity of the fractured area where the crack has grown and where localised plasticity driven fracture process has occurred.

At post-fracture examination, with the aid of SEM micrographs, practically no fracture surface could be observed in this study. This was observed in the broken specimens for Al-foil and LDPE both as individual material layers and when bonded together to form a laminate. The cross-sections are thinned to vanishing thicknesses as an effect of the plastic deformation for both the Al-foil and the LDPE. Under these circumstances, it seems reasonable to assume that the final breaking of the cross-section is obsolete or, at least, that it has very little effect on the fracture process as a whole. Therefore the governing fracture process is based on pure plastic deformation and depending on a strong coupling between the individual layers of the laminates. Additional in situ experimental tests with complementary SEM-micrographs of the plastic strain localisation during the course of stretching are needed. This information will be used to confirm the theoretical conclusions that have been drawn in this work.

6. Conclusions

The fracture or failure process of the freestanding aluminium is a localised plastic deformation and thinning until the cross section vanishes. In the freestanding LDPE localised plastic deformation was not observed. Instead plastic deformation occurs in diffuse regions that surround the crack tip. The plastic region increases to incorporate most of the test specimen at larger loads. In the laminate the aluminium layer behaved similar as a freestanding layer but the LDPE layer switched to localised deformation seemingly forced to do so to comply with the deformation of the aluminium.

The fracture process of thin Al-foil is a localised straining in a strip shaped region that stretches ahead of the crack tip. This is occurring no matter if Al-foil is acting alone or as a member in the laminate. The height/width of this strip is 10–20 \( \mu m \) and the length is in the order of 100–400 \( \mu m \). The fracture process is a continuous plastic deformation forming slip-lines. Furthermore, the process is completed when the thickness in this necked strip region is reduced to zero. Initially the fracture process of the LDPE layer is similar with strains localising to a strip region that more or less seems to coincide with the strip process region in Al-foil layer. However, the thickness does not continuously reduce to zero in the necked strip region for the individual layer of LDPE. Instead the material hardens enough to force material in the vicinity of the necked region to deform instead. This is due to sufficient strain hardening in the material. In the laminate, considerable thinning occurs but then the Al-foil and the LDPE layers detach so that the deforming length increases as the detachment spreads
out from the crack plane. The two materials are at this stage acting by themselves. First the Al-foil breaks completely and finally the LDPE breaks. The thickness is at this point so small in the LDPE that the final fracture surface cannot be observed in the SEM. The main contributions are summarised below:

- A micro-mechanical approach utilising SEM-micrographs, micro-mechanisms, and an analytical expressions motivated the derivation of an equation suitable to calculate the work of failure of freestanding and laminated thin Al-foil.
- A slip-line theory was adopted with a final inclination (1:2) of the failed cross-sections. The theoretical slip-line model is verified for freestanding Al-foil by inspection of SEM micrographs of failed experimental specimens. This theory is not sufficient to explain the governing phenomena and deformation mechanisms of the single LDPE-film since it is deformed significantly more than the Al-foil. LDPE untangle, re-orient and strain-harden during the deformation process.
- The slip-line theory is also applicable on the cross sections in the Al-foil created by the Al-foil laminated with a LDPE layer.
- LEFM (valid approximately when 2a > 15 mm) was used to derive an analytical expression for prediction of the critical load for centre cracked specimens. This equation is accurate both for freestanding Al-foil and a packaging laminate consisting of one-side laminated Al-foil with LDPE. This expression can be used when the plastic region is much smaller than the crack length.

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Anisotropic Elastic-Viscoplastic Properties at Finite Strains of Injection-Moulded Low-Density Polyethylene

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Abstract Injection-moulding is one of the most common manufacturing processes used for polymers. In many applications, the mechanical properties of the product is of great importance. Injection-moulding of thin-walled polymer products tends to leave the polymer structure in a state where the mechanical properties are anisotropic, due to alignment of polymer chains along the melt flow direction. The anisotropic elastic-viscoplastic properties of low-density polyethylene, that has undergone an injection-moulding process, are therefore examined in the present work. Test specimens were punched out from injection-moulded plates and tested in uniaxial tension. Three in-plane material directions were investigated. Because of the small thickness of the plates, only the in-plane properties could be determined. Tensile tests with both monotonic and cyclic loading were performed, and the local strains on the surface of the test specimens were measured using image analysis. True stress vs. true strain diagrams were constructed, and the material response was evaluated using an elastic-viscoplasticity law. The components of the anisotropic compliance matrix were determined together with the direction-specific plastic hardening parameters.

Keywords Polyethylene · Anisotropic · Tensile · Viscoplasticity · Elasticity · LDPE · Injection-moulding · Constitutive behaviour

Introduction

Polyethylene is a polymer of great industrial importance, and the mechanical properties of this polymer are therefore of great interest. Polyethylene normally exhibits a semi-crystalline microstructure. Depending on the details of the polyethylene molecules (i.e. degree of branching, length of branches, etc), different degrees of crystallinity may be attained.

Polyethylene that solidifies in the absence of concurrent deformation of the melt forms a spherulitic microstructure, which is isotropic from a mechanical point of view [1, 2]. The mechanical properties of such isotropic polyethylene have been examined in several experimental studies [e.g. 3–7].

However, there are manufacturing processes that leave the polymer in an anisotropic state. One example is injection-moulding, which is one of the most common manufacturing processes for polymer products. During the injection-moulding process, a polymer melt is forced into a
cavity under high pressure, and the geometry of the cavity determines the shape of the final product. Often one of the dimensions of the product geometry is relatively small compared to the rest of the geometry, and this dimension may therefore be termed the thickness dimension. The mechanical properties of injection-moulded articles are markedly affected by the temperature and velocity fields that melt elements in the cavity are subjected to during the moulding cycle. Layers through the thickness with differing characteristics may therefore be identified.

During the filling stage in the injection-moulding cycle, the bulk (or core) of the polymer cools very little, but next to the mould surfaces, there will be a rapidly cooling thin skin of polymer. This skin layer is normally very thin, has virtually no preferred orientation of the polymer chains, and is normally negligible from a mechanical point of view. However, between the skin and the bulk layers, adjacent layers of polymer melt flow at different velocities and generate shear stresses, which cause the polymer chains to be aligned in the direction of polymer flow. This produces a shear layer, which is highly oriented and is said to be in a state of frozen strain [8–11]. Hence, the injection-moulded material can be expected to be anisotropic and also inhomogenous in the thickness direction.

In the present paper, we investigate the anisotropic mechanical properties of low-density polyethylene (LDPE) plates that have been created through an injection-moulding process. The material is characterized in terms of its anisotropic elastic-plastic behaviour at large strains as well as its strain-rate dependency. The paper is organized as follows: In “Continuum Mechanics Definitions”, we introduce the necessary continuum mechanics entities that are needed to characterize the mechanical properties of the polymer material. In “Materials and Methods”, we provide the details of the experimental procedure and the methods for evaluating the experiments. The results of the experiments are presented in “Results”, and “Discussion” and “Conclusions” contain a discussion of the results and some concluding remarks.

Continuum Mechanics Definitions

Kinematics

We assume the existence of a set of orthogonal basis vectors, \( e_1, e_2, \) and \( e_3 \). Let \( X = X_1 e_1 + X_2 e_2 + X_3 e_3 \) denote the position of a point in the reference (undeformed, unloaded) configuration. Let \( x = x_1 e_1 + x_2 e_2 + x_3 e_3 \) denote the position of the same point in the current (deformed) configuration. The displacement of the point is \( u = x - X \).

In general, the current position and the displacement are functions of \( X \), i.e. \( x = x(X) \) and \( u = u(X) \).

The engineering strain tensor, \( \epsilon \), is defined by

\[
\epsilon = \frac{d\mathbf{u}}{d\mathbf{X}} = \frac{d\mathbf{x}}{d\mathbf{X}} - \mathbf{I},
\]

where \( \mathbf{I} \) is the identity tensor. This measure of strain, defined in equation (1), will be used in the present study. However, when we speak of the ‘engineering’ strain in test specimens below, we mean the local.true engineering strain in the central part of the test specimens, rather than the load-line displacement divided by total initial specimen length that is sometimes termed ‘engineering strain’ in the literature.

Throughout this paper, bold face symbols, such as \( \epsilon \), denote tensors. The components of such tensors are associated with an orthogonal coordinate system and are denoted by \( \epsilon_{ij} \), where \( i, j \in [1, 2, 3] \).

The principal stretches of the deformation are denoted by \( \lambda_1, \lambda_2, \) and \( \lambda_3 \). For an orthotropic material, where a uniaxial tensile test is performed along one of the principal directions of the material, the normal strain components may be expressed as

\[
\epsilon_{11} = \lambda_1 - 1, \quad \epsilon_{22} = \lambda_2 - 1, \quad \epsilon_{33} = \lambda_3 - 1.
\]

The total strain may be decomposed into elastic and plastic parts, i.e.

\[
\epsilon = \epsilon_e + \epsilon_p.
\]

The Jacobian of the deformation, \( J = \lambda_1 \lambda_2 \lambda_3 \), quantifies the volume change of the material, where \( J = 1 \) signifies incompressibility. Plastic deformations may be assumed to be incompressible, and for cases where the plastic strains are far greater then the elastic strains, \( J \approx 1 \) may be assumed. For uniaxial tensile tests performed along one of the principal directions of the material, where the loading has progressed well into the plastic regime, the strain in the thickness direction, i.e. \( \epsilon_{33} \), may therefore be estimated through the relation

\[
\epsilon_{33} = \lambda_3 - 1 \approx \frac{1}{\lambda_1 \lambda_2} = 1 - \frac{1}{(1 + \epsilon_{11})(1 + \epsilon_{22})} - 1.
\]

Stress

We introduce the symmetric (true) Cauchy stress tensor, \( \sigma \), and the non-symmetric first Piola-Kirchhoff (or engineering) stress tensor, \( \mathbf{P} \). In a uniaxial tensile test, where a force, \( F \), is applied to a test specimen, the first Piola-Kirchhoff stress state is given by

\[
P_{11} = \frac{F}{A_0}, \quad \text{all other } P_{ij} = 0,
\]
where $A_0$ is the initial cross-sectional area of the test specimen, and where the force is applied in the $e_1$-direction. The true stress is given by
\[ \sigma_{11} = \frac{F}{A_0\lambda_2\lambda_3} = \frac{P_{11}}{\lambda_2\lambda_3} = \frac{P_{11} \lambda_1}{J}, \] all other $\sigma_{ij} = 0, \text{ (6)}$

where the entity $A_0\lambda_2\lambda_3$ is the current (true) cross-sectional area of the specimen. In the case of incompressible deformation, $J = 1$ holds, implying that $\sigma_{11} = P_{11} \lambda_1$.

### Elastic Material Response

The injection-moulding manufacturing process leaves the polyethylene material in question with three principal directions, and the material may therefore be characterized as orthotropic. The material is taken to be elastic-plastic, and both the elastic and plastic properties may be assumed to be orthotropic.

The elastic behaviour of an orthotropic material requires the determination of 9 independent elasticity constants, and the elastic behaviour can be described by the compliance matrix, according to
\[
\begin{bmatrix}
\frac{1}{E_{11}} & -\frac{v_{12}}{E_{12}} & -\frac{v_{13}}{E_{13}} & 0 & 0 & 0 \\
-\frac{v_{21}}{E_{21}} & \frac{1}{E_{22}} & -\frac{v_{23}}{E_{23}} & 0 & 0 & 0 \\
-\frac{v_{31}}{E_{31}} & -\frac{v_{32}}{E_{32}} & \frac{1}{E_{33}} & 0 & 0 & 0 \\
0 & 0 & 0 & \frac{1}{G_{12}} & 0 & 0 \\
0 & 0 & 0 & 0 & \frac{1}{G_{13}} & 0 \\
0 & 0 & 0 & 0 & 0 & \frac{1}{G_{23}}
\end{bmatrix}
\] \text{ (7)}

where $E_1$, $E_2$, and $E_3$ are the Young’s moduli of the three principal directions, $v_{ij}$, $i, j \in [1, 2, 3]$, are the Poisson’s ratios, and $G_{12}$, $G_{13}$, $G_{23}$ are the shear moduli. Of these 12 parameters inside the $6 \times 6$ matrix, only 9 are independent, owing to the symmetry of the compliance matrix.

The three Young’s moduli and the Poisson’s ratios of the compliance matrix can be determined from three uniaxial tensile tests performed along the three principal directions of the material, and the shear moduli can be determined from three shear tests.

### Plastic Material Response

The uniaxial plastic behaviour may be characterized by the initial yield stress, $\sigma_{y0}$, and a hardening behaviour. The hardening curve describes the evolution of the yield stress, $\sigma_y$, as a function of $\epsilon_p$, the plastic strain in the direction of loading during the uniaxial tensile test. A time-independent yield function, $\sigma_y(\epsilon_p)$, is attained for low loading rates, i.e.

\[ \sigma_y = \sigma_{y0} \left( 1 + H\epsilon_p \right), \] \text{ (8)}

where $H$ is the plastic modulus (normalized by $\sigma_{y0}$).

In general, the uniaxial stress, $\sigma$, required to drive plastic deformation, also depends on the deformation rate, $\dot{\epsilon}_p$, i.e. the material is visco-plastic. In the present study, we assess the visco-plastic behaviour by use of a Johnson-Cook-like viscoplasticity law, namely
\[ \dot{\epsilon}_p = \dot{\epsilon}_0 \left( \exp \left( \frac{\sigma - \sigma_y}{\sigma_r} \right) - 1 \right), \] \text{ (9)}

which is valid for $\sigma > \sigma_y$, where $\dot{\epsilon}_0$ and $\sigma_r$ are material constants, and $\sigma$ is the current stress in the uniaxial tensile test.

For an orthotropic material, each material direction has its own initial yield stress and hardening behaviour.

### Materials and Methods

#### Material

The tested material was a low-density polyethylene (LDPE), which came in plates that had been created through an injection-moulding process, see Fig. 1. The base material utilized was a white-pigmented LDPE with an MFI (Melt Flow Index) of approximately 20 g/10min (190 °C, 2.16 kg; ASTM D1238-04 and ISO 1133:1997). A standard injection-moulding machine, i.e. a horizontal Arburg 470 800-70S hydraulic injection moulding equipment, was used to manufacture the polymeric test plates, following ISO 294-5. The plastic melt, with a temperature of 240 °C, was injected with a volume flow of 20 cm$^3$/s into a tool with a temperature of 40 °C. This was followed by cooling for 11 seconds. The test plates were produced by pure injection-moulding, and the flood gate ensures that the flow front evolves evenly in the test plates, see Fig. 1. The flow direction is denoted by ‘MD’, the direction perpendicular to the flow direction is denoted by ‘CD’, and the thickness direction is denoted by ‘TD’. The plates had a thickness of 0.60 mm.

![Fig. 1 Injection-moulding of test plates](image)
Test specimens, suitable for tensile testing and shaped as dog bones, were punched out mechanically from the polymer plates, see Fig. 2. Three material directions were investigated, i.e. MD, CD and an intermediate direction, $45^\circ$ to MD and CD. The dog bone geometry utilized follows the ISO standard of the test plate. The length of the area where the local strain is measured is 35 mm, the width of the measuring area is 6.0 mm, and the thickness of the specimen is the same as the plate thickness, i.e. 0.60 mm, see Fig. 2.

With regard to indices of stress and strain tensors, we will, throughout this paper, use the index convention $\text{MD} \Leftrightarrow e_1$, $\text{CD} \Leftrightarrow e_2$, and $\text{TD} \Leftrightarrow e_3$, where $e_1$, $e_2$, and $e_3$, constitute a set of orthogonal basis vectors.

Testing Procedure

A standard tensile testing machine was used for the tensile tests, and the testing was performed under displacement control with a prescribed grip separation velocity. All tests were performed at room temperature. The immediate output from the tensile tests was the load vs. time data. Three different loading rates (i.e. grip separation speeds) were used: 0.5 mm/s, 10 mm/s, and 100 mm/s. In terms of strain rate in the central part of the test specimens, where the local strain is measured, these deformation rates correspond approximately to strain rates of 0.0125/s, 0.25/s, and 2.5/s, respectively. The strain rates (as well as the thickness of the plates) were chosen so that they would reflect the deformation processes that applications in the packaging industry (e.g. beverage packages) might be exposed to. For each test condition, three tests were made in order to get an impression of the statistical dispersion in the results. Most of the tests were done using a monotonically increasing displacement, but a few tests were performed with cyclic loading.

Local Strain Measurement

The tests were also video recorded using a standard video camera with a frame rate of 30 frames per second. Furthermore, the local strains on the surface of the test specimens during the tests could be established using image analysis. To facilitate this strain evaluation, lines were drawn on the specimens using a colour pen, as illustrated in Fig. 3(a).

An image analysis toolbox for Matlab was utilized. The analysis method is illustrated in Fig. 3. From the videos, snapshot images are taken with appropriate frequency, so that the straining process is properly resolved. At the highest loading rate, all 30 frames per second had to be used, in order to get a reasonable resolution of the straining process. Markers were defined in a reference image (at zero strain), see Fig. 3(a). After that, the image analysis program tracks the movement of these markers in subsequent images, which represent different degrees of specimen deformation, see Fig. 3(a).

Once the current positions of the markers have been established for a specific image, the displacement of the positions can be plotted vs. the reference positions of the
markers, see Fig. 3(b). A straight line is fitted to the marker displacements, and the slope of this fitted line provides an estimate of the engineering strain in the central part of the specimen, in this case \( \epsilon_{11} = \Delta u_{1}/dX_1 \).

The markers in Fig. 3(a) illustrate the measurement of strain in the loading direction, but markers were used in the same way for measuring the strain in the direction perpendicular to the loading direction. Hence, both the strain in the loading direction and that in the direction perpendicular to loading could be measured.

**Evaluation of Elastic-plastic Behaviour**

In the present study, we only measured the in-plane elastic properties of the injection-moulded LDPE plates. Hence, only \( E_1, E_2, \nu_{12}, \nu_{21} \), and \( G_{12} \) were determined. The first four of these constants were determined from the uniaxial tensile tests in MD and CD. However, \( G_{12} \) was not determined from a shear test but from the tensile tests performed in the 45° direction. The shear modulus \( G_{12} \) may be expressed as

\[
G_{12} = \frac{4}{E_{45}} - \frac{1}{E_1} - \frac{1}{E_2} + \frac{\nu_{12}}{E_2} + \frac{\nu_{21}}{E_1},
\]

where \( E_{45} \) is the Young’s modulus for the 45° direction.

It should be emphasized, that in the derivations above, we have assumed that deformations are confined to the elastic regime and that strains and any possible rotations of the material therefore remain small.

Just as for the elastic properties, the plastic properties of the three in-plane directions were determined in the experiments. Hence, the initial yield stress and plastic modulus for these three directions are denoted by \( \sigma_{y0,1}, H_1, \sigma_{y0,2}, H_2, \sigma_{y0,45}, \) and \( H_{45} \) for the MD, CD and 45° directions, respectively. The associated plastic strains and current yield stresses are denoted by \( \epsilon_{p,11}, \sigma_{y,1}, \epsilon_{p,22}, \sigma_{y,2}, \epsilon_{p,45}, \) and \( \sigma_{y,45} \), respectively.

During a tensile test in MD, for instance, the total strain rate, \( \dot{\epsilon}_{11} \), may then be expressed as

\[
\dot{\epsilon}_{11} = \dot{\epsilon}_{e,11} + \dot{\epsilon}_{p,11} = \frac{\dot{\sigma}_{11}}{E_1} + \dot{\epsilon}_{p,11},
\]

\[
\dot{\epsilon}_{p,11} = \dot{\epsilon}_0 \left( \exp \left( \frac{\sigma_{11} - \sigma_{y,1}(\epsilon_{p,11})}{\sigma_r} \right) - 1 \right)
\cdot \Theta (\sigma_{11} - \sigma_{y,1}(\epsilon_{p,11}))
\]

\[
\epsilon_{p,11} = \int_{t}^{1} \dot{\epsilon}_{p,11} dt.
\]

where \( \Theta (\bullet) \) is the Heaviside step function, which takes on the value 1 for \( (\bullet) > 0 \) and zero otherwise.

In the experiments, a constant deformation rate is applied, which translates to a strain rate in the central section of the specimen that is essentially constant. Hence, \( \dot{\epsilon}_{11} \) may be taken to be constant during the test. Equations (11)–(13) constitute a set of differential equations that can be solved numerically for \( \sigma_{11}(t) \) and \( \epsilon_{p,11}(t) \). The solution also enables the prediction of the elastic-viscoplastic response, i.e. \( \sigma_{11} \) vs. \( \epsilon_{11} \), of the material, which may then be compared with the experimental response. Corresponding equations may be formulated and solved for the CD and 45° directions.

The stress-strain response predicted by equations (11)–(13) is compared to the stress-strain response from the experiments, and in this way, the elastic and plastic properties for the three in-plane material directions are assessed. The estimation of Young’s modulus, Poisson’s ratio, and plastic properties from the experiments is illustrated in Fig. 4. The estimation of \( E_1, \sigma_{y0,1}, \) and \( H_1 \) is provided as an example in Fig. 4(a). In a similar way, estimates of the Young’s moduli and plastic properties of the CD and 45° directions are attained.

The estimation of Poisson’s ratio is illustrated in Fig. 4(b). The evolution of \( \epsilon_{11} \) and \( \epsilon_{22} \) with time is shown. By use of strain data \( \epsilon_{11} \in [0, 0.04] \) (and the corresponding data points for \( \epsilon_{22} \)), straight lines are fitted, as shown in Fig. 4(b). The slopes of these lines are denoted by \( k_1 \)

![Fig. 4](image-url)  
**Fig. 4** Estimation of elastic and plastic properties from tensile tests: (a) Estimation of Young’s modulus, initial yield stress, and hardening through comparison of experimental response (solid line) and model calibration (dashed line); (b) Estimation of Poisson’s ratio through fitting of two straight lines to the two strain paths.
and $k_2$, and the value of Poisson’s ratio is then estimated as

$$v_{12} \approx -\frac{k_2}{k_1}. \quad (14)$$

In a similar way, estimates of $v_{21}$ and $v_{45}$ (the Poisson’s ratio associated with stretching in the 45° direction) are attained.

Results

Force and Strain Data from Experiments

Figure 5 shows the output from a test for MD at a deformation rate of 0.5mm/s. The force data was filtered from noise.

The force data in Fig. 5(a) shows a monotonically increasing force. This was only seen for MD. The force data for the CD and 45° directions showed a decreasing force once the onset of plastic deformation had occurred. The true stress vs. strain relation, however, always showed a monotonically increasing stress, i.e. no true material softening was observed.

In Fig. 5(b) the associated strain data is shown. The strain in the loading direction, $\epsilon_{11}$, shows an almost linear increase with time. The strain in CD, $\epsilon_{22}$, also varies essentially linearly with time, but with a negative slope. The ratio $\epsilon_{22}/\epsilon_{11}$ allows for a determination of the Poisson’s ratio $v_{12}$.

In Fig. 5(b), the estimation of $\epsilon_{33}$, according to equation (4), is also included. In this case, the curve for $\epsilon_{33}$ virtually overlaps the curve for $\epsilon_{22}$. This means that the plastic strains in CD and TD, that are caused by the plastic stretching in MD, are distributed equally to CD and TD.

Strain Response During Monotonic Loading

In Figs. 6, 7, and 8, the outcome from the strain measurements is shown for the three different loading rates, respectively.

Starting with the slowest loading rate in Fig. 6, we note that there is some discrepancy between the strain histories of the different directions. The total elongation of the test specimens is distributed between the thin area of local strain measurement and the wider parts where the specimen is clamped. This distribution appears to differ between the different directions, which causes the observed discrepancies in strain history. The Poisson’s ratios, $v_{12}$ and $v_{21}$, were calculated for each test using equation (14).

The 45° direction was the direction that exhibited the most ductility in the sense that the highest fracture/localization strains were observed for this direction.
This must be attributed to the microstructure of the injection-moulded material. The in-plane strain perpendicular to the 45° direction is denoted by $\epsilon_{45}$.

The central part of the test specimens (where the strain measuring takes place) for the 45° did not undergo a simple stretching during the tests. The material is anisotropic, and when a tensile test is performed in a direction that does not coincide with one of the principal directions of the material, the material will undergo a deformation consisting of both stretching and shearing. However, the shearing was not measured during the present tests.

In Figs. 7 and 8, the outcomes from the tests at the medium and highest loading rates are displayed. Again, there is a systematic discrepancy in the local true strain history between the different directions. The tendency is again that for the CD direction, more of the total strain is distributed to the thin area of measurement compared to the other directions. It may also be noted, that the difference between the strains in the perpendicular directions is more pronounced compared to what was observed at the slowest loading rate.

### Stress-strain Response During Monotonic Loading

Based on the force vs. time and strain vs. time data, the stress-strain response could be established. Since we were not able to measure the strain in TD, we assume incompressibility for all deformations, i.e. also for the elastic deformations. This assumption is necessary to enable the computation of the true stress in the test specimens.

Figure 9 shows the stress-strain results for monotonic loading in MD. The individual coloured lines indicate individual tests. The initial elastic response is similar for the three loading rates, even though a slight non-linearity in the initial response can be identified for the lower loading rates. The initial elastic response is taken to be rate-independent, and Young’s modulus is therefore estimated as the average of all the initial responses from the tests.
The plastic hardening is remarkably linear, and the curves for the three loading rates are virtually parallel. The stress required to drive plastic deformation increases with loading rate, which is clear evidence of the presence of strain-rate effects in the material. The proposed hardening and visco-plasticity laws are able to represent the experimental outcome very well.

Figure 10 shows the outcome for CD. In this direction, there is a pronounced initial non-linearity, at least for the lowest loading rate. Just as in the case for MD, Young’s modulus is taken to be the average for the three loading rates. In addition, the initial yielding behaviour in CD differs from the yielding behaviour of MD. In MD, the hardening is fairly constant after the onset of plastic yielding, whereas in CD, the hardening decreases slightly after onset of plastic yielding, and then after that, the hardening increases again and becomes fairly constant. A consequence of this peculiarity is that the model predictions for CD are relatively inaccurate in the regime where plastic deformation is initiated, but the prediction becomes better once the constant hardening regime is entered.

Figure 11 shows the results for the 45° direction. The 45° direction was the direction that showed the most ductility, and the material could be strained up to almost 200% before failure occurred. The plastic hardening behaviour is fairly linear up to about 100% of straining. After that, the hardening decreases slightly. As a result, the model predictions of the viscoplastic response become increasingly inaccurate at the highest strains.

In Table 1, direction-specific material constants characterizing the elastic and plastic behaviour of the material are tabulated. Hence, the Young’s modulus, the initial yield stress, the plastic modulus, and the visco-plasticity parameters are listed for the three in-plane directions that were tested.

The elastic properties can be given either in terms of stiffness or compliance. In our case, it is convenient to characterize the initial response of the material in terms of material constants associated with the compliance matrix in equation (7). The compliance matrix is a $6 \times 6$ matrix, whose components we denote by $S_{ij}$. Hence, we have $S_{12} = -\nu_{12}/E_1 = S_{21} = -\nu_{21}/E_2$, see equation (7). Thus, we provide estimates of the stiffness values $E_1$, $E_2$, and $G_{12}$ together with the component $S_{12} (= S_{21})$ from the compliance matrix. Based on these estimates, the values of $\nu_{12}$ and $\nu_{21}$ may be calculated if needed.

The elastic constants were estimated as follows. The Young’s modulus $E_1$ was fitted to the initial elastic slope of the test data for MD, as illustrated in Fig. 4. The same procedure was used to establish $E_2$ and $E_{45}$ from the tests in the CD and 45° directions, respectively. For each test in MD and CD, the values of $\nu_{12}$ and $\nu_{21}$ could also be extracted. The associated values $S_{12} = -\nu_{12}/E_1$ and $S_{21} = -\nu_{21}/E_2$ could then be computed, since $E_1$ and $E_2$ had already been

![Fig. 9 Stress strain data from experiments for MD (blue: $\dot{\varepsilon} = 0.0125/s$, red: $\dot{\varepsilon} = 0.25/s$, and green: $\dot{\varepsilon} = 2.5/s$) together with model prediction (dashed black line)](image)

![Fig. 10 Stress strain data from experiments for CD (blue: $\dot{\varepsilon} = 0.0125/s$, red: $\dot{\varepsilon} = 0.25/s$, and green: $\dot{\varepsilon} = 2.5/s$) together with model prediction (dashed black line)](image)

![Fig. 11 Stress strain data from experiments for 45° direction (blue: $\dot{\varepsilon} = 0.0125/s$, red: $\dot{\varepsilon} = 0.25/s$, and green: $\dot{\varepsilon} = 2.5/s$) together with model prediction (dashed black line)](image)
determined. Hence, each test in MD and CD produced one estimate of $S_{12}$ and $S_{21}$, respectively. The final estimate of $S_{12}$ was then taken to be the average of all values of $-\nu_{12}/E_1$ and $-\nu_{21}/E_2$ measured from the tests. Based on the estimates of $E_1$, $E_2$, $E_{45}$, and $S_{12}$, the value of $G_{12}$ was then computed using equation (10). The elastic constants thus attained are listed in Table 2.

It should be noted, that the dispersion in results for $S_{12}$ and $S_{21}$ was relatively high with a standard deviation of about 0.68. The estimation of $S_{12}$ should therefore be regarded as highly approximate. The reason for this uncertainty is most likely that the distance used for measuring the transverse contraction (i.e. the width of the test specimens) was relatively small.

### Failure Strains and Failure Modes

As can be observed from Figs. 6–11, there seems to be a systematic variation in failure strain. For MD, the failure strain increases with loading rate from about 0.6 (on the average) at the lowest loading rate up to 0.9 (on the average) at the highest rate. For CD, the failure strain also increases with loading rate from about 0.75 to 1.0. The 45° direction is by far the most ductile direction. Interestingly, for this direction, the failure strain shows no consistent pattern. That is, the highest loading rate is associated with the lowest failure strain (about 0.7), whereas the lower loading rates are associated with the highest failure strains, ranging from 1.0 up to 1.9.

It should be noted, that the end of the strain history in the tests was not always caused by final failure, but in a few tests, local neck formation in the measuring area prohibited further measurements of the local strain history.

In Fig. 12, some typical fracture surfaces are shown. As can be seen from both Fig. 12(a) and (b), failure in MD and CD occurred more or less perpendicularly to the loading direction and at a well defined failure site. On the other hand, failure in the 45° direction occurred in a zone that was smeared out and approximately oriented 45° to the loading direction. These observations were made at all loading rates.

### Stress-strain Response During Cyclic Loading

In addition to the tensile tests with monotonic loading, we also did two tests with cyclic loading, one for MD and one for CD. The purpose of these cyclic tests was primarily to get at least a qualitative impression of the material performance during unloading. The outcome of these tests is shown in Figs. 13 and 14.

In Fig. 13, the response for MD is shown. Figure 13(a) shows an overview of the stress-strain response for the cyclic load case, and Fig. 13(b) shows a close-up of the unloading process. As can be seen in Fig. 13(b), three unloadings are made at the beginning of the plastic regime. From the hysteresis of the unloading/reloading curves, it is obvious that the material response contains some additional viscosity that is not incorporated in the elastic-viscoplastic model that was used to assess the material behaviour. It is also worth noting, that the general hardening behaviour of the curve with unloading/reloading essentially follows the curves for monotonic loading, although the unloading/reloading process seems to slightly weaken the material response. On the other hand, this discrepancy could also be due to statistical variations in the material response.

In Fig. 14(a) and (b), the result of unloading/reloading for CD is shown, where Fig. 14(b) contains a close-up of the cyclic loading process. Again it is evident that the general hardening behaviour of the curve with unloading essentially follows the same behaviour as the curves with monotonic loading.

### Discussion

In the present experimental study, we have examined the in-plane, anisotropic, elastic-viscoplastic properties of LDPE

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**Table 1** Direction-specific material constants from uniaxial tensile tests

<table>
<thead>
<tr>
<th>Direction</th>
<th>$E$ [MPa]</th>
<th>$\sigma_{00}$ [MPa]</th>
<th>$H$ [-]</th>
<th>$\dot{\varepsilon}_0$ [1/s]</th>
<th>$\sigma_r$ [MPa]</th>
</tr>
</thead>
<tbody>
<tr>
<td>MD</td>
<td>210</td>
<td>12</td>
<td>23</td>
<td>0.1</td>
<td>1.6</td>
</tr>
<tr>
<td>CD</td>
<td>150</td>
<td>9</td>
<td>10</td>
<td>0.05</td>
<td>1.35</td>
</tr>
<tr>
<td>45°</td>
<td>150</td>
<td>9</td>
<td>14</td>
<td>0.1</td>
<td>1.7</td>
</tr>
</tbody>
</table>

**Table 2** Estimated in-plane elastic properties

<table>
<thead>
<tr>
<th>$E_1$ [MPa]</th>
<th>$E_2$ [MPa]</th>
<th>$S_{12}$ [1/GPa]</th>
<th>$G_{12}$ [MPa]</th>
</tr>
</thead>
<tbody>
<tr>
<td>210</td>
<td>150</td>
<td>-3.0</td>
<td>46</td>
</tr>
</tbody>
</table>
Fig. 12 Examples of fracture surfaces for the three test directions; (a) MD, (b) CD, (c) 45°

that has undergone an injection-moulding process. Test specimens were punched out from plates, and because of the small thickness of the plates, only the in-plane properties could be determined.

MD and CD are the principal material directions within the plane of the plates investigated. The stress-strain response of the present polymer differs from what is commonly observed for ductile, low molecular weight materials. The transition from initial elastic to subsequent plastic deformation was smooth, and a specific yield strain was difficult to identify. The initial response was non-linear, especially at the lower loading rates. The fact that the initial response differed between different loading rates is also an indication that there are additional viscous/viscoelastic effects that are not properly accounted for by the elastic-viscoplastic model adopted for the assessment. Furthermore, the non-linear initial response is most likely associated with the deformation of the amorphous sections in combination with rotation of the crystalline segments of the semi-crystalline microstructure of the LDPE material.

The plastic hardening behaviour, on the other hand, was remarkably linear in the stress vs. strain diagrams. Hence, the viscoplastic behaviour could be well described by a linear hardening law for the inviscid plastic response in combination with a Johnson-Cook-like viscoplasticity law. The plastic behaviour of this type of semi-crystalline material is primarily related to the deformation of the crystalline segments of the microstructure. Moderate plastic deformations are associated with dislocation motion within the crystallites, and further plastic deformation causes disruption of the crystalline segments [12–17].

The material response was assessed by the use of an elastic-viscoplastic model. Overall, this model was able to reproduce the material response well, in particular when the material was exposed to monotonic loading. It was evident, though, that there are additional features of the material response that are not fully represented by such a model. This was especially clear from the cyclic loading tests, where it was evident that there were additional viscous/viscoelastic aspects of the material response that are not accurately captured by elastic-viscoplasticity alone.

It is evident from the stress-strain diagrams, that the material is strongly anisotropic. This pertains both to the elastic and plastic behaviour. The Young’s modulus and initial yield stress are about 25% higher for MD compared with CD, and the hardening is also clearly higher for MD. Furthermore, it is clear that the material exhibits a strong strain rate-dependence. When comparing the driving stress at the lowest and highest deformation rates, the difference was about 5MPa for all directions, which amounts to a substantial fraction of the applied stress.

Fig. 13 Stress-strain data for cyclic test (red line) in MD (˙ε = 0.0125/s). The monotonic response (blue lines) is included as a reference. (a) The whole load history; (b) Close-up of unloading/reloading regime
In this study, we consider the homogenized (through the thickness) behaviour of the LDPE plates. One motivation for this is that these types of thin plates are often modelled as shells rather than 3D structures. However, from a material mechanics point of view it is still of interest to characterize the mechanical properties of the different layers of these injection-moulded plates. Hence, further studies are needed, in which the microstructural features of the different layers of the plates are more accurately connected to the macroscopic material behaviour.

At present, we lack detailed knowledge of the microstructure of the injection-moulded material at hand. All three directions tested exhibited a significant amount of ductility. However, it was evident that the 45° direction by far was the most ductile direction. Hence, it seems that the crystalline structure in the material is arranged in such a way that the deformation mechanisms favour plastic deformation along the 45° direction.

It should also be noted, that when tests were made for the 45° direction, the deformation of the thin section of the test specimens did not undergo a simple stretching, as was the case when testing was done in MD and CD. Instead, the material underwent a combined stretching and shearing deformation. This is an outcome that is to be expected when an anisotropic material is stretched in a direction that does not coincide with one of the principal material directions. The strain entity $\varepsilon_{45}$ should therefore, strictly speaking, not be regarded as the normal strain in the 45° direction (except for at small strains) but rather as an equivalent strain quantifying the combined stretching and shearing experienced by the material.

It is of interest to study failure strains as a function of material direction and loading rate. It was noted in the tests, that the statistical dispersion in the failure strains (or in a few cases neck formation strains) is large. Final failure is most likely governed by inhomogeneities and irregularities in the material, and it would require a much larger number of tests if the failure strains were to be determined with statistical significance.

**Conclusions**

The anisotropic elastic-plastic properties of injection-moulded LDPE has been investigated. Tensile tests with both monotonic and cyclic loading were performed, and the local strains on the surface of the test specimens were measured using image analysis. True stress vs. true strain diagrams were constructed, and the material response was evaluated using an elastic-viscoplasticity law. The components of the anisotropic compliance matrix were determined together with the direction-specific plastic hardening parameters.

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**References**

We cannot solve our problems with the same thinking we used when we created them.

Albert Einstein

Advancements in package opening simulations
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Advancements in Package Opening Simulations

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Abstract

The fracture mechanical phenomenon occurring during the opening of a beverage package is rather complex to simulate. Reliable and calibrated numerical material models describing thin layers of packaging materials are needed. Selection of appropriate constitutive models for the continuum material models and how to address the progressive damage modeling in various loading scenarios is also of great importance. The inverse modeling technique combined with video recording of the involved deformation mechanisms is utilized for identification of the material parameters. Large deformation, anisotropic non-linear material behavior, adhesion and fracture mechanics are all identified effects that are needed to be included in the virtual opening model.

The results presented in this paper shows that it is possible to select material models in conjunction with continuum material damage models, adequately predicting the mechanical behavior of failure in thin laminated packaging materials. Already available techniques and functionalities in the commercial finite element software Abaqus are used. Furthermore, accurate descriptions of the included geometrical features are important. Advancements have therefore also been made within the experimental techniques utilizing a combination of μCT-scan, SEM and photoelasticity enabling extraction of geometries and additional information from ordinary experimental tests and broken specimens. Finally, comparison of the experimental opening and the virtual opening, showed a good correlation with the developed finite element modeling technique.

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Keywords: Abaqus; adhesion; constitutive model; opening simulation; progressive damage

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1. Introduction and Background

The macroscopic behavior of the packaging material is today often described by a homogenized material definition in the finite element simulation models. This is due to unavailable experimental results of the mechanical behavior of the individual layers. Furthermore, the adhesion in-between the layers are not included in the FE-model and hence neglected. A much more accurate representation of the reality is obtained if each individual material layer is modeled as a unique layer, represented both with in-plane geometry and an out-of-plane thickness. Therefore, representing the laminated packaging material as individual layers enables more flexible simulation models. This functionality is a pre-requisite when one of layers: thickness or geometry is changed or other load cases are investigated. The mechanical behavior of highly extensible, often denoted ductile polymer films, used in the packaging industry has recently been studied by Jönsson et al. (2013). The polymer materials, consisting of different variants of polyethylene grades, are used in the packaging material structure at Tetra Pak® today.

A significant re-orientation of the polymer chains and a substantial strain-hardening occurred during the deformation process in the experimental uniaxial tensile tests. The latter effect is very important and has to be accounted for in the numerical material modeling approach. The simulations were solved in the general finite element software Abaqus version 6.13 (2013). In this work a continuum damage modeling (CDM) approach was used for each individual material layer to represent the fracture mechanisms. CDM which is attractive in macro scale applications, thus solving the engineering problems, was chosen in this study due to the computational efficiency.

A damage criterion consisting of two functionalities: initiation of damage and evolution of damage was suitable for modeling the ductile fracture behavior, cf. Andreasson et al. (2012). During the numerical analysis it has been assumed that the polymer materials are anisotropic, homogenous through the thickness, independent of strain rate and independent of temperature to ease the material parameters identification. Similar material modeling approach was used for the less extensible aluminum foil, also present in the laminated packaging material structure. A package with a post applied opening device is included for illustration in Fig. 1. This is an example of an application that is simulated numerically in this paper to show the maturity of the simulation strategy.

During the opening process four topics/mechanisms are important to control, understand and accurately quantify:

- Mechanical material behavior - stretching of the membrane, all packaging material layers
- Progressive damage material behavior - cutting of the membrane, all packaging material layers
- Adhesion - traction law between the individual packaging material layers, all packaging material layers
- Contact/interaction - friction between the cutter/membrane and between the frame/cutter/cap, all included parts

![Fig. 1. Tetra Prisma Aseptic® package with a post applied screw cap opening to the left, the packaging material structure to the right.](image-url)
The membrane, that is cut through, during the opening process consists of a packaging material structure that is shown in Fig. 2. The packaging material membrane consists of four different layers: decor polymer, laminate polymer, aluminum foil and inside polymer. Furthermore, in the finite element simulation model is the cutter part in the opening device together with the membrane included with a dense element mesh as shown in Fig. 2.

Fig. 2. Material structure of the membrane to the left and the finite element model of the membrane and the cutter to the right.

2. Identification of material parameters

An accurate continuum material model is fundamental when incorporating fracture mechanical behavior in the material model in the FE-simulations. The material properties of each material layer were determined by performing experimental uniaxial tensile tests. Individual thin films were tested, consisting of the same materials and produced with similar manufacturing process as the layers in the laminated packaging material, i.e. each polymer film represented a unique layer in the packaging material. Numerical material model parameters were identified with the inverse modeling approach complemented with the photoelastic effect, cf. Jönsson et al. (2013). This was easily adopted and possible to accomplish due to the thin transparent polymer film. Accounting for a significant strain hardening in the polymer layers is important in these highly extensible polymer films.

The results from the calibrated continuum material models used in the virtual tensile tests replicating the experimental tensile test are presented in Fig. 3. A very good fit was possible to obtain when strain-hardening was included in the two different polymer material models. Most often the material model is later on used beyond the validity of the calibration. The stress state can also be more complex and for instance include a cyclic behavior with a combined loading/un-loading scenario. In the presented simulation model the primarily focus is on the monotonic loaded mechanical behavior including progressive damage behavior.

Fig. 3. Comparison of the virtual and the experimental tensile test response graphs with the corresponding deformation to the right.
A result graph from an experimental test, cf. Fig. 3., is a combination of geometrical effects, micro-mechanical mechanism and continuum material behavior. It is very important to be aware of this mixture of effects and hence try to extract the “real” and true material behavior from the specific experimental test-setup that has been performed. The material model used in the finite element software should typically not include the geometrical effects and geometrical shape effects. During the material parameter identification process, hence the inverse modeling phase should this be accounted for. A video recording or even better a Digital Image Correlation (DIC) could be used when solving the inverse problem. Otherwise the risk is large of finding a non-unique solution to the inverse problem that is not the most accurate one. The benefit of using a video that capture the deformation sequence correlated with the experimental data is also to be able to understand the involved mechanism during the experimental tensile test. Furthermore, visualization of the deformation sequence together with the data is possible afterwards.

3. Two virtual simulation models of the package opening

Solving opening simulations in an explicit framework has both advantages and disadvantages. Contact algorithms are much more mature and easily adopted with the general contact framework now available in commercial FE-codes. Progressive fracture modeling is also a conditionally unstable event and is most often impossible to solve in an implicit code today as the authors understanding. Explicit codes was originally developed and customized for rapid and dynamic events like a car crash or drop test. The opening process on the other hand can be done rapidly but the challenge is when it is done very slowly by the customer. Small elements used to resolve a high resolution have an additional cost in an explicit code. Thus decreasing the time increment, and extends the time to solve if the total time event is rather long in reality. Numerical tricks have to be utilized such as semi-automatic mass-scaling to find a good balance between the simulation time and the experimental quasi-static steady state results.

It is very important to account for each individual material layers thickness with their respective mechanical behavior both in respect of continuum behavior and fracture mechanical responses. The packaging material layers are all extrusion coated, the polymers are applied as molten layers, and laminated with aluminum foil at high temperature. This results in an adhesion value that also has to be accounted for and needs to be included in the simulation model. The advantages of modeling the layers individually are that a single layer can be changed and different levels of adhesion can be defined between the layers.

Two different FE-simulation models were developed, one FE-model that described the membrane by using the internal composite layup within one shell element definition. The second FE-model, where the different layers were described by individual shell elements connected with cohesive contact. The membrane was modeled with a circular geometry build up with a dense mesh composed by first order shell elements with reduced integration, S3R. The FE-model can be seen in Fig. 2. The reason for using three node elements, instead of four node elements, was because it was possible to create a more stochastic mesh with equally sized three node elements for the geometry. This method with stochastically distributed elements enabled an arbitrary fracture path in the progressive damage behavior, instead of a predetermined fracture path. The cutter was controlled by a kinematic coupling to a reference point with the aim to mimic the experimental test movement. A vertical displacement and rotation around the central axis of the cutter was assigned. The edge of the membrane was locked in all degrees of freedom and thus no consideration was taken to the flexibility of the paperboard edge connected to the membrane at the outer circumference.

Both simulation models used the same contact definition, general contact with a penalty friction definition. Furthermore, both models used double precision in the submission command to the solver, because of the large amount of increments needed. Five integration points was used through the thickness direction of the shell element.

3.1. One shell element model – full adhesion level

The finite simulation model where the membrane was described using the composite layup definition was very stable, i.e. it was insensitive to changes of e.g. the material properties and the boundary conditions. Furthermore, no damping or stabilization was needed in order for the model to converge and it was possible to use high level of mass scaling and still obtain an accurate and reliable numerical solution. The disadvantage of the composite layup model is that it is only possible to simulate full adhesion between the packaging material layers. Furthermore, it is not
possible for layers that are not initially adjacent to each other to interact, i.e., it is not possible for the inside polymer layer to interact with the laminate polymer layer. But still it is possible to have different continuum material and damage modeling approach assigned to each material definition.

3.2. Four shell elements model — different adhesion levels

In this simulation model all material layers in the membrane, in total four layers, were represented by four separate shell elements with different cohesive contact in-between. Numerical stabilization had to be introduced when solving the model with four individual shell elements in order to get a converged solution. The linear bulk viscosity parameter was changed. The FE-model was sensitive to the level of mass scaling. However, the model was dependant on the total time of the simulation, i.e., the deformation speed. It is important to emphasize that the effect from the deformation speed, mass scaling and stabilization on the simulation results are not independent.

The advantages of the model with four shell elements are that it is possible to include different levels of adhesion between the layers as well as controlling the level of adhesion at different sections between two layers. It is also possible to model geometry of the layers in a more accurate way compared with the composite layup model. Furthermore, it is possible for all layers to interact with each other, i.e., when the aluminum foil breaks it is possible for the inside polymer layer to come into contact with the laminate polymer layer.

4. Findings and Conclusions

In this work and by Pagani et al. (2012) it has been shown that it is now possible in an opening finite element simulation, both in respect of hardware and software, to numerically model each packaging material layer in the membrane as individual layers connected with a cohesive contact. The advantage with this approach is that it is easy to change single layers mechanical properties, thickness or geometrical shape. The FE-simulations accurately describe and are able to accurately predict the opening procedure when the four shell elements model is used.

The final results from the experimental and virtual opening process are shown in Fig. 4. The simulation results mimic the experimental behavior satisfactory for both models. It is definitely possible to predict the opening force level and the overall behavior. The four shell elements FE-model has a cohesive behavior implemented between the shell elements, defined as a contact interaction. The simulation model with full adhesion, one shell element, overestimates the cutting force which is reasonable due to full interaction between the packaging material layers. Including adhesion in the virtual model is a better representation of the reality and hence is the force response curve behavior and peak force value predicting the experimental results very good.

![Fig. 4. Comparison of the reaction force vs. angle from the experimental test performed in Modena (Italy) and the two virtual opening models.](image)
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Appendix A. Abaqus keywords used in the two finite element simulation models

A short description and summary of the important and specific utilized keywords extracted from the Abaqus *.inp-file that was solved in the numerical opening simulation.

* ** MATERIAL DEFINITIONS **
  *Material, name=Aluminium_foil
  *Density
  *Elastic
  *Plastic
  *Damage Initiation, criterion=DUCTILE

  *Material, name=Polymer_film
  *Density
  *Elastic
  *Plastic
  *Damage Initiation, criterion=DUCTILE

* ** COHESIVE BEHAVIOR **
  *Damage Initiation, criterion=MAXS
  *Damage Evolution, type=ENERGY, mixed mode behavior=POWER LAW, power=1.

* ** STEP: Virtual package opening **
  *Step, name=Cutting_through_membrane, nlgeom=YES
  *Dynamic, Explicit
  *Bulk Viscosity

* ** Mass Scaling: Semi-Automatic **
  *Whole Model
  *Variable Mass Scaling, dt=4e-08, type=below min, frequency=100

* ** INTERACTIONS - general contact **
  *Contact, op=NEW
  *Contact Inclusions, ALL EXTERIOR

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research profiles where the data of the publications can be found

ResearchGate

Google Scholar

https://orcid.org/0000-0001-8325-9226
ABSTRACT

The final goal of this PhD-work is an efficient and user-friendly finite element modelling strategy targeting an industrial available package opening application. In order to reach this goal, different experimental mechanical and fracture mechanical tests were continuously refined to characterize the studied materials. Furthermore, the governing deformation mechanisms and mechanical properties involved in the opening sequence were quantified with full field experimental techniques to extract the intrinsic material response. An identification process to calibrate the material model parameters with inverse modelling analysis is proposed. Constitutive models, based on the experimental results for the two continuum materials, aluminium and polymer materials, and how to address the progressive damage modelling have been concerned in this work. The results and methods considered are general and can be applied in other industries where polymer and metal material are present.

This work has shown that it is possible to select constitutive material models in conjunction with continuum material damage models, adequately predicting the mechanical behaviour in thin laminated packaging materials. Finally, with a slight modification of already available techniques and functionalities in a commercial general-purpose finite element software, it was possible to build a simulation model replicating the physical behaviour of an opening device. A comparison of the results between the experimental opening and the virtual opening model showed a good correlation.

The advantage with the developed modelling approach is that it is possible to modify the material composition of the laminate. Individual material layers can be altered, and the mechanical properties, thickness or geometrical shape can be changed. Furthermore, the model is flexible and a new opening design with a different geometry and load case can easily be implemented and changed in the simulation model. Therefore, this type of simulation model is prepared to simulate sustainable materials in packages and will be a useful tool for decision support early in the concept selection in technology and development projects.