

RAPID CRACK PROPAGATION IN POLYMER MULTI - LAYER SYSTEMS

by

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ABSTRACT

This work is concerned with the analysis of rapid crack propagation (RCP) in Polymethylmethacrylate (PMMA), Polycarbonate (PC), Polyvinylchloride (PVC) and two-layer PMMA/PC and PMMA/PVC systems. The experiments were performed employing three different specimen configurations. RCPs in PMMA, PC and PVC were measured using single edge notch tensile (SENT), glass-PC SENT and glass-PVC SENT specimens, respectively. Duplex SENT specimens were employed for the recording of RCPs in the PMMA/PC and PMMA/PVC systems. The testing was carried out at room temperature. Two methods were used to monitor crack length histories; a fast photography and a grid line technique. It was observed that remarkably constant crack velocities were associated with uniform fracture surfaces. RCPs in the PMMA/PC duplex specimens were characterized by a higher crack speed in the PMMA than in the PC section. On the contrary, the crack propagated at a higher speed in the PVC than in the PMMA part of the PMMA/PVC duplex specimens, and crack branching always occurred in the PVC section. The difficulties in dynamic fracture calculations are associated with inertia and stress wave propagation effects. Because of these, values of the stress intensity factor (K^{dyn}) and energy release rate (G^{dyn}) in dynamic fracture events cannot, in general, be derived from the crack length, loading and geometry as in the static case. The presence of dynamic effects makes analytical solutions scarce and possible only for infinite geometry problems. Thus, an indirect method of characterizing RCP properties was used in this study. The method relies on the recorded crack length history and boundary conditions, which are incorporated in a dynamic Finite Element (FE) code to generate the crack resistance (G_{ID}). The program can analyse symmetrical bodies of isotropic linear elastic material. Because of symmetry, only half of the sample was modelled and the crack path becomes a boundary. The crack was propagated by the sequential release of crack path nodes. The calculation of RCP properties was based on the recorded crack length history, the magnitude of the breaking load and single dynamic modulus. It was assumed that as long as the crack

propagates, the equality between the crack driving force (G^{dyn}) and the material crack resistance is satisfied. It is generally accepted that a fracture surface roughness is an indication of the energy absorbed in brittle fracture, and hence the variation of G_{ID} with crack length should correlate with changes in the surface roughness. The numerical simulation of constant speed RCPs generated highly scattered G_{ID} data. Very large variations of the computed G_{ID} with crack length do not correspond to fracture surface appearances. Geometry dependent and multivalued crack resistance data with respect to the crack speed, as calculated by elastodynamic FE code, casts doubt on the uniqueness of G_{ID} . In this work, attempts were made to overcome these difficulties by exploring the concept that the anomalies arise from large local strains around the rapidly moving crack tip, resulting in the crack 'seeing' a low local modulus. Measurements of the strain histories along the crack path were performed in order to support the idea. While the measured peak strains were nearly constant with crack length, the numerical simulation based on single constant modulus produced increasing peak strains over a considerable length of the advancing crack. Unfortunately, the parameters describing the behaviour of the tested materials near the propagating crack tip are uncertain. Therefore, it was decided to approximate local non-linear effects by a local strip along the prospective crack path. Both the thin strip layer and the surrounding material were taken as isotropic and linearly elastic. A very good fit between the measured and calculated peak strains line was obtained when a low modulus was used for a strip along the crack path. The local strip, low modulus, model generated a larger amount of the kinetic energy in the sample than the single modulus approach. The crack resistance was reduced, compared to that of the single constant modulus analysis. G_{ID} data were nearly independent of the crack length, speed and the sample size. It was shown that this local modulus concept may be successfully applied to the analysis of RCP in the duplex specimen configuration. Values of G_{ID} and local modulus, obtained from the FE analysis of the PMMA SENT samples, were in a very good agreement with those generated by a two spring-single mass model.

Key words: PMMA, PC, PVC, RCP, multi-layer systems, dynamic crack resistance, local low modulus

To
Mirna and Davor

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CONTENTS:

	page
Title page.....	1
Abstract.....	2
Acknowledgements.....	5
Contents.....	6
List of tables.....	10
List of Figures.....	11
Notations.....	15
Abbreviations.....	18
Chapter 1: General background	
1.1 Introduction.....	19
1.2 Some conventions of fracture mechanics.....	19
1.2.1 Static definitions of fracture.....	20
1.2.2 Dynamic fracture definitions.....	23
1.3 Stress wave propagation.....	29
1.4 Note on the uniqueness of the crack resistance vs crack speed..	34
Chapter 2: RCP related experimental methods	
2.1 Introduction.....	41
2.2 Materials and background	
2.2.1 Materials.....	42
2.2.2 Brief review of experimental methods on RCP in polymers.....	43
2.3 Methods of recording dynamic fracture behaviour	
2.3.1 Direct toughness measurement.....	45

2.3.2 Crack speed measurement.....	47
2.4 Review of some experimental works related to RCP in multi-layer systems.....	49
2.4.1 Embrittlement of ductile materials.....	51
2.4.2 Reported works on duplex specimen.....	55

Chapter 3: Experiment on duplex PMMA/PC SENT specimen

3.1 Introduction.....	61
3.2 Duplex specimen details.....	62
3.3 Preliminary tests.....	65
3.4 Setup and testing.....	67
3.5 Crack speed in the duplex specimen	
3.5.1 Note on the crack propagation behaviour.....	73
3.5.2 Duplex specimen crack speed results.....	75

Chapter 4: Experiments on PMMA, PC and PVC

4.1 Introduction.....	85
4.2 SENT specimen details	
4.2.1 PMMA SENT sample.....	86
4.2.2 Glass-duplex SENT sample.....	87
4.2.3 Duplex PMMA/PVC SENT sample.....	90
4.3 Setup and crack speed measurement.....	90
4.4 Results	
4.4.1 PMMA SENT results.....	93
4.4.2 Glass-PC SENT results.....	99
4.4.3 Glass-PVC SENT results.....	105
4.4.4 Duplex PMMA/PVC SENT results.....	106

Chapter 5: Numerical analysis of experimental results

5.1 Introduction.....	108
5.2 FE modelling of RCP	
5.2.1 General basis.....	109
5.2.2 Simulation of the propagating crack.....	110
5.2.3 Details of the FE program used in this study.....	112
5.2.4 Choice of elastic modulus.....	115
5.3 Elastic generation mode analysis of experimental results	
5.3.1 General features.....	118
5.3.2 Elastic analysis of duplex PMMA/PC SENT specimen.....	121
5.3.3 Elastic analysis of PMMA SENT specimen.....	129
5.3.4 Elastic analysis of glass-PC SENT specimen.....	136

Chapter 6: 'Local modulus' model

6.1 Introduction.....	142
6.2 Discussion of the elastic analysis results	
6.2.1 The nature of G_{ID}	143
6.2.2 The sources of possible errors in the elastic analysis.....	152
6.3 Local strip analysis	
6.3.1 Introduction.....	159
6.3.2 Measurement of the strain along the crack path during RCP.....	160
6.3.3 Comparison between the experimental and numerical ϵ_{yy} vs t results	163
6.3.4 Local strip with low modulus.....	167
6.3.5 PMMA SENT summary.....	176
6.3.6 Local low modulus analysis of glass-PC SENT sample.....	180
6.3.7 Local low modulus analysis of duplex PMMA/PC SENT sample...	182

Summary and recommendations.....	185
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References.....	188
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Appendices :

Appendix A1: Stress wave propagation in non-linear material.....	202
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Appendix A2: Mass-Spring modelling.....	204
--	------------

Appendix A3: Influence of the local strip parameters on G_{ID} results	209
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Appendix A4: Summary of G_{ID} results of PC	211
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List of tables :

- 2.2.1 - Static and dynamic material properties at room temperature
- 3.2.1 - The summary of the DS dimensions
- 3.5.2.1 - Selected results of the test on duplex PMMA/PC SENT samples
- 6.3.2.1 - Strain measurement - samples dimensions and SGs positions
- 6.3.4.1 - The peak strains vs G_{ID} predictions
- 6.3.7.1 - Local parameters of the PMMA and PC, for the analysis of of the duplex PMMA/PC SENT sample TMN10-2

List of figures :

- 2.4.1 - Typical tensile load-displacement curve for ductile fracture
- 2.4.2 - Typical tensile load-displacement curve for brittle fracture
- 2.4.2.1 - Schematic presentation of the condition for surface embrittlement (after Rolland[83])
- 3.2.1 - Duplex SENT specimen configuration
- 3.4.1 - Patterns of triggering lines - 'V' prenotch
- 3.4.2 - Patterns of triggering lines - 'V' + sharp prenotch
- 3.4.3 - Sample with black ink
- 3.4.4 - Streak image of a crack propagating at a constant speed
- 3.4.5 - Masked sample
- 3.4.6 - Photographic configuration
- 3.5.2.1 - The streak mode photograph of completely fractured sample (TMN10-2)
- 3.5.2.2 - The streak mode photograph of completely fractured sample
- 3.5.2.3 - The streak mode photograph of a sample with the crack arrest at the interface
- 3.5.2.4 - The streak mode photograph of a sample with the crack arrest at the interface (TBW6-9)
- 3.5.2.5 - The crack arrest at the PMMA/PC interface for a short period
- 3.5.2.6 - The crack arrest at the PMMA/PC interface for a long period
- 3.5.2.7 - The crack arrest at the PMMA/PC interface for a long period
- 3.5.2.8 - Instantaneous crack arrest in the PC part of the duplex sample
- 3.5.2.9 - Permanent crack arrest in the PC part of the duplex sample
- 3.5.2.10- The crack front shape and the streak photography
- 3.5.2.11- Fracture surfaces of completely broken samples
- 3.5.2.12- Typical load-displacement curves of DS with varying initial notch
- 4.2.1.1 - SENT specimen configuration
- 4.2.2.1 - Glass-duplex sample configuration
- 4.2.2.2 - Glass-PVC SENT samples with various D/H

- 4.3.1 - Timing line configuration, 4 line pattern on PMMA SENT sample
- 4.3.2 - Timing line configuration, 8 line pattern on PMMA/PVC SENT sample
- 4.3.3 - Timing line configuration, 12 line pattern on glass-duplex SENT sample
- 4.4.1.1 - Voltage vs time record from 4 line pattern, test PM4-4
- 4.4.1.2 - Voltage vs time record from 8 line pattern, test PM5-1
- 4.4.1.3 - PMMA fracture surface, $\dot{a} \approx 230$ m/s
- 4.4.1.4 - PMMA fracture surface, $\dot{a} \approx 200$ m/s
- 4.4.1.5 - PMMA fracture surface, $\dot{a} \approx 380$ m/s
- 4.4.1.6 - Voltage vs time record of non-uniform speed RCP (test PM9-4)
- 4.4.1.7 - Voltage vs time record of non-uniform speed RCP (test PM6-4)
- 4.4.2.1 - Gross roughening of PC fracture surface, $\dot{a} \approx 450$ m/s
- 4.4.2.2 - Irregular voltage vs time record from 8 line pattern
- 4.4.2.3 - Irregular voltage vs time record from 12 line pattern
- 4.4.2.4 - Uniform fracture surface of PC, $\dot{a} = 437.5$ m/s
- 4.4.2.5 - Multiple crack branching in PC, $\dot{a} \approx 550$ m/s
- 4.4.2.6 - Voltage vs time record from 4 line pattern
- 4.4.3.1 - Crack branching in PVC
- 4.4.4.1 - PMMA/PVC sample, typical fracture surface appearances
- 5.2.4.1 - Schematic response of a VE material to a slow loading and fast unloading
- 5.3.1.1 - Finite Element mesh
- 5.3.2.1 - Crack length increment vs time; various initiation cases; test TBW6-9
- 5.3.2.2 - Crack resistance vs crack length increment; various initiation cases; test TBW6-9
- 5.3.2.3 - Crack length increment vs time; various initiation cases; test TMN10-4
- 5.3.2.4 - Crack resistance vs crack length increment; various initiation cases; test TMN10-4
- 5.3.2.5 - Crack length increment vs time; various initiation cases; test TBN10-6
- 5.3.2.6 - Crack resistance vs crack length increment; various initiation cases; test TBN10-6
- 5.3.2.7 - Crack resistance vs crack length increment; influence of the duration of the crack hesitation at the interface on G_{ID} of PC; test TMN10-2

- 5.3.2.8 - Energy vs time; test TMN6-3
- 5.3.2.9 - Energy vs time; test TMN6-5
- 5.3.3.1 - Crack length increment vs time; various interpolations of experimental data
- 5.3.3.2 - Crack resistance vs crack length increment; influence of various interpolations
- 5.3.3.3 - Crack length increment vs time; test PM4-4
- 5.3.3.4 - Crack length increment vs time; test PM5-1
- 5.3.3.5 - Crack resistance vs crack length increment; influence of various D/H ratios
- 5.3.3.6 - Crack resistance vs crack length increment; influence of various D/H ratios
- 5.3.3.7 - Crack resistance vs crack length increment; influence of the mesh size
- 5.3.4.1 - Glass-PC specimen and the crack initiation
- 5.3.4.2 - Crack resistance vs crack length increment; influence of the crack behaviour in the glass-slide on G_{ID} of PC
- 5.3.4.3 - Crack resistance vs crack length increment; influence of T_{AC} on G_{ID} of PC
- 5.3.4.4 - Crack resistance vs crack length increment; influence of the crack speed
- 5.3.4.5 - Crack resistance vs crack length increment; influence of the specimen size
- 6.2.1.1 - Normalized energies vs crack length increment, test PM4-2, $H=40$ mm, influence of the specimen height
- 6.2.1.2 - Normalized energies vs crack length increment, test PM7-1, $H=80$ mm, influence of the specimen height
- 6.2.1.3 - Fixed-grip uniform displacement test, influence on the crack speed
- 6.2.1.4 - The configuration of the partially clamped PMMA SENT sample
- 6.2.1.5 - 'Partially clamped' boundary conditions test, influence on the crack speed
- 6.2.1.6 - 'Partially clamped' boundary conditions test, influence on the crack speed
- 6.2.2.1 - Crack driving force vs crack length increment, influence of the conditions of the loaded boundary during RCP, on G^{dyn} results
- 6.2.2.2 - Crack resistance vs crack length increment; influence of various single moduli

- 6.3.2.1 - Strain measurement configuration
- 6.3.2.2 - Strain vs time records from two strain gauges, test SG1-1/4
- 6.3.2.3 - Strain vs time records from two strain gauges, test PCSG-2
- 6.3.3.1 - Comparison between experimental and numerical strain results, test SG1-1/4
- 6.3.3.2 - Comparison between experimental and numerical strain results, test PCSG-2
- 6.3.4.1 - Local strip model configuration
- 6.3.4.2 - Peak strain vs crack length increment, test SG1-1/4
- 6.3.4.3 - Comparison between G_{ID} results from different FE analyses, test SG1-1/4
- 6.3.4.4 - Peak strain vs crack length increment, test PCSG-2
- 6.3.4.5 - Comparison between G_{ID} results from different FE analyses, test PCSG-2
- 6.3.4.6 - Energy components in SG1-1/4 sample
- 6.3.4.7 - Energy components in PCSG-2 sample
- 6.3.5.1 - Strain from SG1 vs total crack length, results from three specimens
- 6.3.5.2 - Local modulus vs crack speed
- 6.3.5.3 - Crack resistance vs crack speed, results of PMMA SENT samples
- 6.3.6.1 - Crack resistance vs crack length increment, test PC4-5, influence of E_I on G_{ID}
- 6.3.7.1 - Crack resistance vs crack length increment, constant modulus vs local strip analyses;
influence of the duration of the interface crack hesitation on local strip results,
test TMN10-2
- A1.1 - Stress-strain curve for bilinear material
- A1.2 - Waves propagation through the bar with bilinear stress-strain relationship
- A2.1 - The mass-spring model
- A2.2 - $1/EG_0$ vs $\dot{a}^2$, experimental results
- A2.3 - $\bar{C}$ vs D/H
- A3.1 - Influence of E_I on G_{ID} results; $l=0.433$ mm, test SG1-1/4
- A3.2 - Influence of l on G_{ID} results; $E_I=0.1$ GPa, test SG1-1/4
- A4.1 - Crack resistance vs crack speed, PC results

Notations :

a	-	crack length
$\dot{a}$	-	crack speed
B	-	specimen thickness
C	-	compliance
$\bar{C}$	-	characteristic crack speed
C_L	-	longitudinal stress wave speed
C_T	-	shear stress wave speed
C_R	-	Rayleigh wave speed
D	-	specimen width
d	-	increment of crack growth between nodes
E	-	Young's modulus
E_s	-	static Young's modulus
E_d	-	dynamic Young's modulus
F_h	-	holding back force
FY_i	-	loads on upper boundary nodes
G	-	energy release rate
G^{st}	-	statically determined energy release rate
G^{dyn}	-	dynamically determined energy release rate
G_{IC}	-	critical energy release rate (mode I, plane-strain)
G_{ID}	-	dynamic crack resistance
G_{IA}	-	minimum value of G_{ID} (dynamic crack resistance at arrest)
H	-	specimen height
K	-	stress intensity factor
K^{st}	-	statically determined stress intensity factor

K_{dyn}	-	dynamically determined stress intensity factor
K_{IC}	-	fracture toughness (mode I, plane-strain)
K_{ID}	-	dynamic fracture toughness
K_{Ia}	-	statically determined stress intensity factor at arrest
K_{IA}	-	(or K_{Im}) minimum value of K_{ID} (crack arrest toughness)
K_M	-	machine stiffness
P	-	externally applied load
P_b	-	externally applied load prior to crack initiation
R	-	crack resistance
t	-	time
Δt	-	time-step size
T	-	temperature
T_{AC}	-	time interval between moment of triggering and breaking of the first crack gauge
u	-	loaded boundary displacement
U_s	-	strain energy
U_k	-	kinetic energy
w	-	sample width
W	-	external work

Greek letters :

ϵ	–	strain
ϵ_{yy}	–	strain in y direction
ϵ_{zz}	–	strain in z direction
γ_0	–	surface energy
γ^{dyn}	–	dynamic surface energy
γ	–	shear strain
σ	–	stress
σ_{yy}	–	stress in y direction
σ_{zz}	–	stress in z direction
σ_{ys}	–	yield stress
ν	–	Poisson's ratio
ρ	–	mass density

Abbreviations :

ABS -	acrylonitrile-butadiene-styrene
CHV -	cross head velocity (of a tensile loading machine)
COD -	crack opening displacement
CTOD -	crack tip opening displacement
CT -	compact tension
DCB -	double cantilever beam
DS -	duplex specimen
HIPS -	high impact polystyrene
PC -	polycarbonate
PE -	polyethylene
PMMA -	polymethylmethacrylate
PVC -	polyvinylchloride
QS -	quasistatic
RCP -	rapid crack propagation
RT -	room temperature
SAN -	styrene-acrylonitrile
SEN -	single edge notch
SENB -	single edge notch bending
SENT -	single edge notch tensile
TPB -	three point bending
VE -	viscoelastic

CHAPTER 1

GENERAL BACKGROUND

1.1 Introduction

The purpose of this work is to investigate Rapid Crack Propagation (RCP) in polymer multi-layer systems. In order to understand the nature of RCP, the first chapter reviews briefly the general fundamentals in fracture mechanics theory. A comprehensive description of fracture in polymers from the macroscopic point of view (continuum mechanics) and the microscopic one (atomic and molecular theories) can be found in several text-books such as Williams[84] and Kinloch and Young [85]. An extensive review of dynamic fracture problems is recently given by Freund[90]. However, section 1.2 considers some conventions of fracture mechanics (static and dynamic fracture), while the stress wave propagation is summarized in section 1.3. The note on the uniqueness of the crack resistance (G_{II}) vs crack speed ($\dot{a}$) is given in section 1.4.

1.2 Some conventions of fracture mechanics

The macroscopic theories of fracture assume the existence of cracks and of other imperfections which may act as fracture nuclei. The size of those 'initial' cracks is assumed to be sufficiently large, compared to the characteristic dimensions of the microstructure, to justify the use of the continuum mechanics tools. The material is considered to be a homogeneous continuum with certain idealized properties.

From a macroscopic point of view, the fracture of solids may be classified into two broad categories: brittle fracture and ductile fracture. Ductile fracture is usually associated with large deformations, high rate of energy dissipation and slow crack growth. Brittle fracture is a low-energy failure which takes place in a catastrophic manner. Brittle fracture is generally associated with high crack velocity. This work concentrates on the analysis of rapidly moving crack in polymers.

1.2.1 Static definitions of fracture

In general, there are four forms of the crack behaviour: initiation, propagation, branching and arrest. The analysis of RCP is concentrated on a 'short time' propagating event. For easier understanding of the fracture parameters related to a moving crack, the fundamental definitions of stationary crack parameters are briefly reviewed in this section.

Since the pioneering work by Griffith[21] in which he devised an energy-rate analysis of equilibrium of cracks in brittle materials, known as the energy balance approach to fracture, conventional linear elastic fracture mechanics has become well established. Fracture processes are defined as the formation of new surfaces in a medium in a thermodynamically irreversible manner (continuum mechanics approach). According to the mode of the crack loading there are three modes of fracture: opening, sliding and tearing, notated as mode I, II, III respectively. This work concentrates on the opening mode I only.

The conditions in a body containing a crack can be defined by two approaches: local and global. In the local approach consideration is given to the singularity in the elastic stress field in the vicinity of the crack tip. This singularity is characterized by a parameter called the stress intensity factor (K). The intensity of the stress field ahead of the crack tip is 'controlled' by K

and all the near-tip stress components are proportional to K in the following form:

$$\sigma_{ij} = \frac{K}{\sqrt{2\pi r}} f_{ij}(\phi) \quad 1.2.1.1$$

σ_{ij} - cartesian stress components; r, ϕ - polar coordinates around the crack tip; $f_{ij}(\phi)$ - trigonometric function.

For an infinite plate with a stationary central crack of length $2a$, with uniformly applied normal stress σ at infinity, K is given as:

$$K = \sigma \sqrt{\pi a} \quad 1.2.1.2$$

For a stationary crack in systems of finite dimensions, other length factors appear and K is expressed as:

$$K = \sigma \sqrt{\pi a} Y\left(\frac{a}{w}\right) \quad 1.2.1.3$$

w - sample width; $Y(a/w)$ - geometry correction factor.

The global approach considers the total energy balance in the sample. Suppose that a crack of length a grows an infinitesimal distance δa , and let δU denote the change in the total energy during the extension of the crack. The ratio $\delta U / \delta A$, where $A = B \delta a$ is the crack surface area and B is the thickness of the sample, can be interpreted as a force (per unit length) tending to cause crack extension. This amount of energy that would be released during this first increment of crack growth is called G - elastic (strain) energy release rate or crack driving force, and is expressed by:

$$G = \frac{dW}{dA} - \frac{dU_s}{dA} \quad 1.2.1.4$$

dA - unit area of crack surface; W - external work; U_s - strain energy.

For stationary crack, G can be calculated from the loading conditions, sample geometry and stiffness:

$$G = \frac{P^2}{2B} \frac{dC}{da} = \frac{1}{2B} \left(\frac{u}{C} \right)^2 \frac{dC}{da} = \frac{U_s}{B} \frac{1}{C} \frac{dC}{da} \quad 1.2.1.5$$

P - external force; B - sample thickness; C - specimen's compliance; u - applied displacement.

The global crack driving force can be interpreted in terms of the local crack tip stress field, i.e. G can be related to K :

$$G = K^2/E' \quad 1.2.1.6$$

$$E' = E \quad \text{for plane stress}$$

$$E' = E/(1-\nu^2) \quad \text{for plane strain.}$$

The stress solution (equation 1.2.1.1) predicts infinite value for the stresses at the crack tip. This cannot occur in practice, and a 'real' material always yields plastically near the crack tip. Under plane stress conditions ($\sigma_{zz}=0$) there is a large plastic zone where normal stress σ_{yy} reaches the uniaxial yield stress σ_{ys} , while under plane strain conditions ($\epsilon_{zz}=0$) the plastic zone is of different shape and much smaller due to triaxial stress constraint. This constraint also causes σ_{yy} to reach an average of $1.68 \sigma_{ys}$ (Broek[74]).

The crack extension may occur when K reaches its critical value at initiation or likewise, when G is equal to the energy required for the crack to initiate. These critical K and G are defined as the fracture toughness and the crack resistance of the material, respectively.

Under plane strain conditions the crack reaches the critical initiation conditions at minimum values of K and G . Therefore, material properties defined by these unique and minimum (worst case) values; the critical fracture toughness - K_{IC} and the critical crack resistance - G_{IC} , are related to plane strain conditions (the subscript 'I' denotes for mode I).

In practice one is generally concerned with preventing rather than promoting the fracture. Therefore, most of the research has been concentrated primarily on the determination of G_{IC} or K_{IC} for the different materials. There are nevertheless a number of important cases where information on the behaviour of moving cracks is extremely useful, e.g. information about the efficient methods of stopping a running crack or producing controlled rapid fracture.

1.2.2 Dynamic fracture definitions

The difficulties in analysing dynamic fracture problems are associated with the effects of inertia and of the stress wave propagation. Due to these dynamic effects the momentary values of the stress intensity factor and energy release rate in dynamic fracture events cannot be derived from the crack length, loading and geometry as in static cases. Therefore, systematic dynamic analysis has to be employed. This section considers the definitions of some dynamic fracture parameters and difficulties in their characterization by analytical methods.

In dynamic fracture, the stress intensity factor around the moving crack tip is defined as for a stationary crack:

$$K^{dyn} = \lim_{r \rightarrow 0} (\sigma_{yy}(r,0) \sqrt{2\pi r}) \quad 1.2.2.1$$

while G^{dyn} is defined as the total dynamic energy release rate. Inertia effects and kinetic energy must be taken into account in the calculation of those quantities. The energy balance gives:

$$G^{\text{dyn}} = \frac{1}{B \frac{da}{dt}} \left(\frac{dW}{dt} - \frac{dU_s}{dt} - \frac{dU_k}{dt} \right) = \frac{dW}{dA} - \frac{dU_s}{dA} - \frac{dU_k}{dA} \quad 1.2.2.2$$

(U_k - kinetic energy).

Analytical solutions concerning RCP problems can be classified roughly as:

- a) approximate solutions - these can be regarded as static energy balance results from which a kinetic energy correction is subtracted, and methods where the results of exact dynamic solutions are approximated to finite specimen geometries;
- b) exact dynamic solutions - these divide into the solutions which rely on the steady state assumptions (it is assumed that the near tip stress field is independent of time and does not develop with increasing crack length) and transient solutions without such assumptions (time dependence is not removed). Exact dynamic solutions have been obtained for idealized cases of dynamic fracture - infinite plate and infinite strip cases.

An extended review of both dynamic fracture solutions is given by Crouch[86], and only a few 'classical' exact solutions will be mentioned here.

Yoffe[51] was one of the first to attempt a dynamic analysis of a moving crack. The analysis relied on the steady-state assumption. This state was achieved by considering a non-realistic case of constant length crack running at constant speed in an infinite plate (the crack was opening at one end and simultaneously closing at the other). Broberg[60] obtained the most outstanding transient solution for a central crack initiating from zero length and expanding symmetrically at constant speed in an infinite plate under constant remote stress. He considered a hypothetical material of zero toughness and derived a limiting crack speed equal to the Rayleigh wave speed (C_R) from the fact that the crack tip stress singularity disappeared at C_R .

(section 1.3). In [64], Broberg analysed a more realistic case in which the stress singularity at the tip of the moving crack was removed and stresses limited to a certain maximum value. He argued that the material elastic constants and consequently the stress waves velocities, in highly strained crack tip region, differed appreciably from those in the unstrained medium. So, he speculated that the limiting crack speed could be lower than C_R in the unstrained material. Broberg[67] related G^{dyn} to the statically determined energy release rate - G^{st} , through the crack velocity dependent correction function $g(\dot{a})$. The result is an elastic solution of the form:

$$G^{dyn} = G^{st} g(\dot{a}) \quad 1.2.2.3$$

Freund[72-a] analysed a pre-existing semi-infinite crack that started to propagate at a constant velocity in an infinite plate under a remote uniform tensile stress. As a result, the relationship between K^{dyn} and the statically determined stress intensity factor - K^{st} was derived (equation 1.2.2.4). It takes the same form as the relationship depicted in equation 1.2.2.3.

$$K^{dyn} = K^{st} k(\dot{a}) \quad 1.2.2.4$$

Both $k(\dot{a})$ and $g(\dot{a})$ are unity at zero crack velocity and reduce to zero at C_R . In these constant speed solutions, the dynamic corrections ($g(\dot{a})$, $k(\dot{a})$) depend only on the crack speed.

The relationship between K^{dyn} and G^{dyn} (Freund[72-a]), retains the form as in the static case:

$$G^{dyn} = \frac{(1 - v^2)}{E} (K^{dyn})^2 f(\dot{a}) \quad 1.2.2.5$$

$f(\dot{a})$ is a function of the crack speed $\dot{a}$, the dilatational (longitudinal, irrotational) wave speed C_L and the distortional (equivoluminal, shear) wave speed C_T (section 1.3). $f(\dot{a})$ is equal to one at zero crack speed and tends to infinity at C_R . However, the dependence of K^{dyn} on speed ensures G^{dyn} goes to zero at C_R .

Freund[72-a, second part] also simulated a crack travelling at variable speed. He argued ([72-a, 72-b]) that the constant crack speed results (equations 1.2.2.3 - 5) may be applied to any arbitrary crack length history, since this may be modelled as incremental steps of constant crack speed. It was demonstrated by Nilsson[74] that the form of the near tip stress distribution was independent of the crack acceleration or specimen geometry and depended only on instantaneous crack speed. Hence, Freund's relation (equation 1.2.2.5) became universally valid for any linear elastic dynamic fracture problem.

An infinitely long strip of finite height, fracturing along its length, was analysed by Nilsson[72]. The steady-state was achieved by assuming that the crack has been propagating for such a long time, that the stress field became insensitive to the crack length (time) due to the constraint imposed by the loaded boundaries. For the clamped boundary case (zero lateral contraction) his analysis resulted in G^{dyn} equal to the static result at all crack velocities, while K^{dyn} was found to be dependent on the crack speed and equal to zero at C_R . G^{dyn} remained non-zero at C_R and both G^{dyn} and K^{dyn} showed insensitivity to boundary stress wave reflections (a steady-state condition involving repeated stress arrivals at the crack tip was rapidly established).

The cases above are for Mode I tensile fracture and the following points can be made regarding these analytical works:

a) The equation 1.2.2.3 is valid for any crack length history in an infinite plate. This demonstrates that the dynamic correction is principally influenced by near crack tip instantaneous conditions. Equations 1.2.2.3-4 remain valid in finite specimens until a stress wave reflected from a boundary has returned to the crack tip. These analytical results can be used for testing dynamic numerical models (e.g. finite element (FE) model) as well as for quantitative evaluation of the finite geometries before the first arrival of reflected stress waves to the crack tip.

b) The theoretical crack speed upper limit is the Rayleigh wave speed (C_R). Even in a material with zero toughness, the crack cannot propagate faster than C_R , because a disturbance cannot travel along a surface faster than the surface stress-wave. An increasing speed imposes a shorter time within which near crack tip (necessary) displacement level can be attained. Consequently, inertia effects consume the released strain energy which cannot flow fast enough to reach the crack tip and be dissipated there (as the crack speed approaches C_R).

In reality, the terminal crack speed may differ considerably from the theoretical prediction (C_R for unstrained material). The material properties and consequently the stress wave speeds, in the region near the moving crack tip, could be significantly different from those in the unstrained or moderately strained material away from the tip. While the high strain rate increases the wave velocities, large strains and high temperature in the vicinity of the crack tip have the opposite effect - lower stress wave speeds with the consequence that the energy flow to the tip is reduced (Broberg[64], [73]). Ravi-Chandar and Knauss [84-b, 85] speculated that the process of generating micro-cracks and their subsequent coalescence at the crack tip requires an interaction between those micro-fractures which necessitates a certain time. Thus, they concluded that cracks cannot propagate with near wave velocity as long as multiple micro fractures play a role. These factors could serve as a support for the experimentally observed terminal crack speed lower than C_R .

G^{dyn} and K^{dyn} are 'state variables' describing the potential of the system to feed energy to the moving crack tip. They are dependent on the specimen geometry, mechanical properties and the mode of loading, and thus they are not material properties. In order to generate the material resistance to RCP the only valid approach is to define the material properties related to the crack propagation phase. Other approaches attempt to approximate RCP properties ignoring the crack propagation phase:

- a) the dynamic crack initiation approach simulates the behaviour of the running crack under constant stress by the dynamic initiation conditions of a stationary crack under increasing stress conditions,
- b) the static crack arrest approach ignores the propagation phase and the statically determined stress intensity factor at crack arrest (K_{IA}) is assumed to be a measure of the material crack arrest toughness (K_{IA}).

The material properties related to the crack propagation phase are defined as: dynamic fracture toughness (K_{ID}) and the dynamic crack resistance (G_{ID} or R). K_{ID} describes the stress field close to the moving crack tip while G_{ID} defines the energy absorbed by the propagating crack. The dynamic approach to RCP gives direct consideration to the crack propagation and arrest. Within the confines of elastodynamic behaviour, the condition for unstable crack propagation takes the form:

$$G^{dyn} = G_{ID} \quad (\text{or equivalently} \quad K^{dyn} = K_{ID}) \quad 1.2.2.6,$$

where, G_{ID} (K_{ID}) is temperature dependent function of the crack speed. G^{dyn} is the crack driving force (dynamic energy release rate) and K^{dyn} is dynamic stress intensity factor. The crack arrest occurs at the position and time when the crack driving force cannot supply the minimal energy required to sustain propagation, or when K^{dyn} becomes less than the minimum value of K_{ID} for all greater times (Kanninen[85]):

$$G^{dyn} < G_{IA} \quad (\text{or equivalently} \quad K^{dyn} < K_{IA}) \quad 1.2.2.7,$$

where, G_{IA} (K_{IA}) denotes the minimum value of G_{ID} (K_{ID}) with respect to the crack speed.

1.3 Stress wave propagation

In the theory of elasticity the body is considered to be in equilibrium under applied forces, and the elastic deformations are assumed to have instantly reached their static values. This is sufficiently accurate for problems in which the time between the application of a force and the setting up of effective equilibrium is short compared with the times in which the observations are made. However, events when the body is subjected to forces which are applied for only very short periods of time or are changing rapidly, must be considered in terms of the propagation of stress waves. The deformations produced by those forces, and so the information about the instantaneous stress distribution, travel through the body in the form of stress waves. RCP events are accompanied by stress wave propagation. The region around the crack tip, subjected to a high tensile stresses (mode I), is suddenly relaxed and a compressive stress wave is generated as soon as the crack starts to propagate. The sample can be regarded as unbounded before the first stress wave reflected from the sample boundary has reached the crack tip. During RCP, stress waves are emitted continuously from the wake of the moving crack tip (from the loading region in impact), and are reflected from the sample boundaries. This section briefly summarizes general concepts of the stress waves propagation. Comprehensive review of the stress waves in solids can be found in several text books such as Kolsky[63] and Johnson[72].

The stress waves can be transmitted through the body - body waves, or on its surface - surface waves. In extended isotropic solids two types of elastic waves may be propagated: dilatation waves and distortional waves, travelling at velocities C_L and C_T respectively. When the body has a free surface or a surface boundary exists between two solids, elastic waves may be propagated along the surface of a solid. These surface waves are known as Rayleigh waves, and the disturbances associated with them decay exponentially with depth.

The particle motion associated with a dilatational wave is along the direction of wave propagation. This wave, also called irrotational, consists of longitudinal waves and involves no rotation (although, both dilatation and distortion are involved). The particle movement caused by a distortional wave is perpendicular to the direction of wave propagation (the motion is transverse and parallel to the wave front). This wave is also called equivoluminal or shear wave and involves no dilatation (change of shape at constant volume).

When the body wave of either type impinges on a boundary between two media, both reflection and refraction occur. Generally, four separate waves are generated: a wave of each type is reflected and a wave of each type is refracted. There can be no refracted waves at a free surface (ideally in vacuum). Thus an incident dilatation wave is reflected from a free surface at the angle of incidence and with amplitude less than that of the incident wave. A distortion wave is reflected as well, and the sum of the energies of two reflected waves is equal to the energy of the incident wave. At normal incidence of dilatation wave there is no reflected distortion wave, and the amplitude of the reflected dilatation wave is equal to that of the incident wave. Similarly if a distortion wave is incident on a free boundary both a distortion and a dilatation wave are generally reflected. At normal incidence only the distortion wave is reflected with an amplitude equal to that of the incident wave.

If an elastic wave of either type impinges the boundary between two solids, in general, two waves (dilatational and distortional wave) are reflected back, and two of these are refracted into the second medium. At normal incidence, waves of the same type as the actual incident wave are generated, and if the product of the mass density and wave speed is equal for the two media, no wave will be reflected.

When the 'mechanical impedance' - ρC of the second medium is higher than that of the first, the reflected dilatation wave is of the same sign as that of the incident dilatational wave (ρ is the mass density in unstrained state and C is dilatation wave velocity). When the medium on the other side of the surface has the lower 'mechanical impedance', the reflected dilatation wave

changes its sign from tension to compression or vice-versa.

A sudden increase in the medium cross-section causes reflection of the same sign while abrupt decrease in cross-section results in reflection with changed sign.

The velocity of the wave propagation should be distinguished from the particles (mass) velocities. According to the laws of motion particles accelerate from a stationary position and their speed is a function of the magnitude of the stress at that point ($\sigma = \rho C du/dt$). Particles can move in an opposite direction to the stress wave (tension wave).

If two stress waves traveling in opposite directions overlap, the resulting stress and the resulting velocity of particles in the portion of overlapping are obtained by superposition. Assume that a compression wave is traveling along the bar and a tension wave of the same length and with the same stress magnitude is moving in opposite direction. In the portion of the bar where the two waves are superposed, the resulting stress is equal to zero, while at the same time, the resulting velocity of particles is doubled (this portion may be considered as a free boundary). Contrary, if two identical waves, moving toward one another, overlap, there will be double resulting stress and zero resulting speed in the portion the waves are superposed (this portion may be considered as a fixed boundary).

The distortion wave speed is geometry independent and its velocity - C_T is given as:

$$C_T = \sqrt{\frac{G}{\rho_0}} \quad 1.3.1$$

G - shear modulus; ρ_0 – density in unstrained state.

The dilatation wave speed - C_L , depends on the ratio between wavelength and the sample dimensions.

Along a 'thin rod', with the cross sectional dimensions less than the wavelength and with no displacement constrained to zero, the dilatation wave propagates at:

$$C_L = \sqrt{\frac{E}{\rho_0}} \quad 1.3.2$$

E - Young's modulus.

In a case of 'thin plate' whose thickness is smaller than the wavelength (including the 'thin rod' case with strain constrained to zero in one lateral direction), C_L is given as:

$$C_L = \sqrt{\frac{E}{\rho_0 (1-\nu^2)}} \quad 1.3.3$$

ν - Poisson's ratio.

For the 'thin plate' whose transverse deformation is constrained to zero (or for the 'thin rod' with two lateral constraints) the dilatation wave (longitudinal) velocity is:

$$C_L = \sqrt{\frac{E(1-\nu)}{\rho_0(1+\nu)(1-2\nu)}} \quad 1.3.4.$$

So, when the wavelength becomes comparable with lateral dimensions of the solid the velocity of longitudinal waves depends on the wavelength. At very short wavelengths (an unbounded solid) this velocity reduces to the speed of the surface waves.

The surface Rayleigh-wave speed depends on material properties, and it is given as:

$$C_R = f(\nu) C_T \quad 1.3.5$$

(e.g. for $\nu=0.34$, $f(\nu)=0.9415$; etc).

Rayleigh waves consist of elliptical movement of the particles close to a free surface.

For the cases in which material non-linearity exists, the dilatation waves can be propagated at different speeds depending on the stress-strain relation (appendix A1). From the simple one dimensional equilibrium equation:

$$\frac{\partial \sigma}{\partial x} = \rho \frac{\partial^2 u}{\partial t^2} \quad 1.3.6$$

$$\text{since } \epsilon = \frac{\partial u}{\partial x} \text{ then } \frac{\partial \epsilon}{\partial x} = \frac{\partial^2 u}{\partial x^2} \text{ so :}$$

$$\frac{\partial^2 u}{\partial t^2} = \left(\frac{1}{\rho} \frac{\partial \sigma}{\partial \epsilon} \right) \frac{\partial^2 u}{\partial x^2} \quad 1.3.7$$

thus the speed of the wave propagation along the x-axes is:

$$C = \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\epsilon}} \quad 1.3.8$$

For the linear case $\partial \sigma / \partial \epsilon = E$ as before.

The large strains around the crack tip would lead to locally small values of $\partial \sigma / \partial \epsilon$ causing much smaller wave speed. This observation is shown to be very important in the analysis of RCP (chapter 6).

1.4 Note on the uniqueness of the crack resistance vs crack speed

This work considers RCP in polymer 'double-layer' systems for cases where the crack path is perpendicular to the interface between two materials. The principal aim is to characterize material properties related to RCP in each part of the system. In addition to the analysis of RCP in the samples consisting of two different materials, each part of the 'double-layer' specimen will be separately tested and analysed as well. Throughout this work, it is assumed that as long as the crack propagates the equality between the crack driving force and the material resistance exists (equation 1.2.2.6), i.e. energy is conserved. If G_{ID} (K_{ID}) is to be a material property (unique G_{ID} vs $\dot{a}$ relationship) then the material under observation, regardless of whether it is a part of the multi-layer system or it is a 'single layer' specimen, must experience the same G_{ID} at a given crack speed.

In practice, the G_{ID} vs $\dot{a}$ relation is extremely important since it defines the conditions for crack arrest. Generally, the region around a rapidly propagating crack tip is influenced by high strain rate, large strain, adiabatic heating etc. These factors may have diverse effects on G_{ID} and any guess of the variation of the crack resistance with the crack speed is very uncertain. In strain rate sensitive materials at low crack speeds, the crack resistance might be expected to decrease with speed due to an increase in the yield stress at higher strain rate (fracture toughness is determined by the plastic zone size which is an inverse function of the square of yield stress). Large strain, unlike high strain rate, lowers the stress wave speed and reduces the energy flow into the crack tip. The adiabatic heating is dominant at higher crack speeds. Thus, the crack resistance might be expected to increase with speed due to a decrease in the yield stress.

The dynamic fracture toughness (K_{ID}) versus crack speed ($\dot{a}$) relations of brittle polymers (Epoxies, Homalite-100, Polymethylmethacrylate etc.) and steels (4340 steel, A533B steel

etc.), obtained through extensive fracture testing performed by different authors, were shown to have the same pattern: roughly a crack speed independent region of K_{ID} at low speeds, followed by drastic increase of K_{ID} to a 'near vertical stem' at, so called, terminal crack speed. These relations also revealed the influence of the different geometries on the fracture toughness behaviour. Both, the terminal crack speed and the speed-independent K_{ID} have been reported as significantly dependent on the specimen geometry. The question of uniqueness of K_{ID} vs $\dot{a}$ arose as a consequence of a large data scatter in published relations.

Kobayashi and Mall [78-a] compared K_{ID} results of Homalite-100 and of Araldite B obtained by themselves and by other teams of investigators. They concluded that the errors generated through the use of static near-field solution to compute the dynamic stress intensity factors (from recorded dynamic isochromatics or dynamic caustic - section 2.3.1) might be contributory factors to the scatter of results. Throughout his work, Kobayashi considered K_{ID} as identical to K^{dyn} during crack propagation. He [78-b, 84] favoured a unique K_{ID} vs $\dot{a}$ curve and supported the minimum running crack toughness (K_{Im}) concept in explaining crack arrest. He concluded that "there is mounting evidence to support the existence of a unique K_{ID} vs $\dot{a}$ relation" of polymers and steels. In his recent paper, Kobayashi [86] attributed the data scatter to the experimental errors involved in crack velocity measurements and to the size of the crack tip region used for data reduction in the presence of stress waves (see also Rosakis and Zehnder [85]). He ended by saying that K_{ID} vs $\dot{a}$ relation may possess a unique K_{Im} in case of large samples in the absence of stress wave effects. Such unique, speed independent region of K_{ID} , was not observed in small or dynamically loaded specimens. Dally et al [85] reviewed experimental results, presented by different research groups, on the K_{ID} vs $\dot{a}$ relationship and discussed the possible reasons for the apparent differences in those results. Several sources of errors, when using either the method of caustics or photoelasticity (section 2.3.1), were mentioned: errors due to the three dimensional state of stresses at the crack tip, errors due to the

presence of non-singular terms and the errors due to inaccurate measurement of the crack speed. Dally noted that these non-singular terms necessitate a higher order representation (higher order terms were usually neglected in the crack tip field analysis). He claimed that the differences in K_{ID} vs $\dot{a}$ curves were not large in view of the possible sources of errors. Nevertheless, Dally believed that if the methods of the crack speed recording were improved, and if dynamic effects were included in the computation of the crack tip features, a unique K_{ID} vs $\dot{a}$ relationship could be established at least for the crack speed independent K_{ID} region. Rosakis et al [83] concluded that solids fracturing in a ductile manner seem to be well described by a unique K_{ID} vs $\dot{a}$ relation while non-unique behaviour has been noted in brittle polymers and high strength steel. Rosakis and Zehnder [85] reported experimental evidence as well as theoretical support for a unique dependence of K_{ID} on the crack tip speed in metals that fracture with a locally ductile mechanism of hole growth. They speculated that large errors in velocity records, caused by stress wave effects, were a main source of uncertainty in the interpretation of K_{ID} results. The consequence of uncertainties in the crack speed record, in the presence of stress wave reflections, were demonstrated by Dahlberg et al [80]. A large scatter in data was observed in the smaller single-edge notched tensile (SENT) specimens where a stress wave would be expected to cause variations in $\dot{a}$, unlike in the larger specimens. This seems unlikely to provide an explanation for geometrical differences. Rosakis and Ravi-Chandar [86] explored the regions where two-dimensional approximations would be acceptable to extract K_{ID} data from either caustics or photoelasticity measurements. The local caustics measurements have been usually made at very small distances from the crack tip, to insure a K-dominant two-dimensional asymptotic field, implying negligible higher order term effects. Because this technique measures the stress field much nearer the crack tip than in photoelasticity, it is important to allow for dynamic effects in the calibration of K_{ID} (Crouch[86]). Rosakis and Ravi-Chandar [86] demonstrated that plane stress conditions prevail at the distances from the crack tip greater than half the sample thickness, while no significant plane strain region was detected. They showed that the large errors in K_{ID} were

generated when the measurements were made at distances less than half the specimen thickness, where the plane-stress results are not valid. However, the errors due to the neglect of the higher order terms in the far-tip region could be substantial (Dally et al [85]). Very recently, Suzuki[91] reported the work on the influence of specimen thickness on caustic K_{ID} results. Dynamic fracture toughness data were measured simultaneously by the caustic and the Crack Tip Opening Displacement (CTOD) methods (section 2.3.1). Suzuki[91] investigated the size of the region where the two methods yield similar results. This caustic applicable region was found to be bounded by two distinct fields where large differences between the caustic and CTOD results appeared: i) near the crack tip region where the caustic K_{ID} results were influenced by the 3D state of the stresses and, ii) far tip region with substantial effects of the higher order terms on the caustic data. As the specimen thickness increased, the size of the caustic applicable region decreased. Suzuki[91] showed that the method of caustic cannot be accurately applied for the PMMA samples with a thickness larger than about 12 mm. Apart from investigating the compatibility of the static and dynamic approaches to RCP, Kanninen[85] considered nonlinear aspects of dynamic fracture mechanics. He pointed out that the question of the uniqueness of K_{ID} could be resolved only by resort to nonlinear dynamic treatment, and that local nonlinear effects must have been responsible for non-uniqueness of K_{ID} . He also demonstrated that geometry dependence and much of the speed dependence of K_{ID} were removed in a viscoplastic treatment of experimental RCP data. Linear Elastic Fracture Mechanics (LEFM) may be greatly invalidated in dynamic fracture, even by modest levels of local non-linearities (Broberg[73]). Crouch[86] investigated RCP in polymeric materials: PMMA, PC, PVC and Araldite. Apart from a significant degree of scatter in generated G_{ID} data, he reported systematic differences in results obtained from different geometries. He favoured the existence of a basically unique curve of fracture resistance versus crack speed, and concluded that the choice of a single constant elastic modulus was the largest single influence on the results. Crouch pointed out that the use of the idealized infinite plate

solutions in the calibration of K_{ID} to the caustic diameter could be a contributory factor to the apparent differences in K_{ID} values between different geometries.

Data on the dynamic stress intensity factor as a function of crack velocity have been obtained by Kalthoff and his colleagues [80, 83]. Most of the data have been generated by means of the shadow optical method of caustics. They reported a large scatter and a geometry dependence in K_{ID} vs $\dot{a}$ results. Kalthoff[83] speculated that a minimum fracture energy requirement may exist at any crack speed, but a material may absorb more if more is available. Shockey et al [83] measured moving crack tip features simultaneously by caustics (section 2.3.1) and thermocouples. They reported a difference between K^{dyn} measured by Caustics and K_{ID} calculated from thermocouple readings. K^{dyn} and K_{ID} characterize the available energy and the energy absorbed by the crack, respectively. They speculated that discrepancies arose from the failure of a global energy balance. That is, the crack tip energy balance is not valid at high crack speeds since some of the available energy can be radiated away by stress waves and only a part of the energy is therefore absorbed at the crack tip. K^{dyn} was derived directly from the linear elastic stress-strain field around the crack path using the method of caustics. K_{ID} was determined from the temperatures measured on the specimen surface near the running crack by assuming that all the plastic energy was dissipated as heat, and applying one-dimensional heat-flow equation for an infinite plane to a three-dimensional specimen with finite boundaries. However, uncertainties involved in the determination of both K^{dyn} and K_{ID} are too high for the failure of the global energy balance to be accepted. These uncertainties may arise from the interaction of stress waves with the crack tip, from a negligence of higher order terms and non-linear effects in the K^{dyn} calibration, from the application of the heat-flow equation which cannot describe the measured temperature-time profile (see Shockey et al [83]). Further, the scatter in both K^{dyn} and K_{ID} data at various crack speeds was at least as large as the reported difference between K^{dyn} and K_{ID} . A viscoelastic material may leave a remnant 'lens' behind the true crack tip position since recovery takes a finite time. Hence, it is preferred that low rate dependence materials are analysed by the method of caustic (Beinert and Kalthoff [81]).

Using a magnetic pulse method of crack face loading of Homalite samples, Ravi-Chandar and Knauss [84-b, 85] reported a constant crack velocity even though the stress intensity factor changed markedly with crack length (K_{dyn} was determined by caustics). This suggested a non-unique K_{ID} vs $\dot{a}$ relation. The effects of specimen size and geometry on K_{ID} data were attributed to the interaction of stress waves with the crack tip. The crack speed was found to depend only on initial tip conditions and loading rate. It was largely independent on subsequent stress conditions, provided sudden changes in the stress intensity factor did not occur due to rapid wave interaction (see chapter 6). They argued that the crack speed was controlled by the distribution of microcracks propagating ahead of the main crack tip. This distribution depends on the distribution at initiation as set by the loading rate. They further claimed that propagation in this non-continuum model necessarily involved interaction between microcrack tips. This 'communication' set an effective stress wave speed below the continuum value. It was noted (Knauss[84], Ravi-Chandar and Knauss [84-a]) that the finite time required to establish a steady state stress distribution for caustic measurements may be a significant contribution to K_{ID} variation, presented as K_{ID} bands at each crack speed.

The dependence of the crack driving force on the crack velocity gradient was reported by Takahashi and Arakawa [87, 90], Williams[87], Kalthoff[88], etc. For the same momentary crack speed the dynamic fracture toughness was found to be larger during deceleration than acceleration, which is reverse of results obtained by Kobayashi and Dally [77]. Contrary, the results published by Rosakis and Zehnder [85], Crouch[86], Dahlberg et al [80], showed no K_{ID} (G_{ID}) dependence on the crack acceleration/deceleration. The crack itself, as distinct from the surrounding sample material, would be expected to behave in an inertialess manner (Kanninen[85]).

The work done by Eftis and Liebowitz [76] can be regarded as a definite representative of non-unique G_{ID} vs $\dot{a}$ stand. Contrary to Griffith's constant surface energy γ_0 , he introduced thermodynamic surface energy density γ^{dyn} as a function of crack speed and surface kinetic energy, and not a function of higher order derivatives of crack velocity. Due to the presence of the load dependent surface kinetic energy, fracture energy could not be regarded as a material property.

CHAPTER 2

RCP RELATED EXPERIMENTAL METHODS

2.1 Introduction

The first chapter briefly reviewed the complex nature of the analysis of rapid crack propagation (RCP). Some of the confusing trends concerning G_{ID} (K_{ID}) vs $\dot{a}$ relation were also mentioned. However, in this work, it is assumed that G_{ID} is a unique RCP property of a given material, and that equation 1.2.2.6 is valid as long as the crack propagates.

Dynamic crack resistance has to be evaluated from experimental data either through 'direct' or 'indirect' methods. This evaluation necessitates a knowledge of the mechanical properties of materials. The description of the materials used in the experimental part of this study is given in section 2.2. Section 2.3 considers methods of the toughness and crack speed measurement.

A brief review of works on surface embrittlement and RCP in duplex specimen (DS) geometry is given in section 2.4.

2.2 Materials and background

2.2.1 Materials

The materials tested in this work were commercially available grades supplied in a sheet form. The bulk of the experimental data were obtained on polymethylmethacrylate (PMMA) and polycarbonate (PC). Limited tests were performed on polyvinylchloride (PVC). The PMMA was ICI Perspex, the PC was General Electric Lexan and the PVC was ICI Darvic.

The mechanical properties of the materials involved in this study are both strain rate and temperature dependent. These properties were established experimentally by many authors (e.g. Crouch[86] measured the mechanical properties of the PMMA, PC and PVC). The appropriate elastic modulus for use in the calculation of dynamic energy release rate in viscoelastic materials is considered in section 5.2.4.

Crouch[86] obtained lower bound (static modulus) estimates from conventional three-point bend specimen measurements incorporating a shear correction. Tensile measurements produced a modulus value less than 5% higher than the bending modulus value. The dynamic modulus data were obtained by using direct contact ultrasonic modulus measurements. These modulus data were used throughout this work. The mechanical properties of the materials involved in the present study are summarized in table 2.2.1.

Property	PMMA	PC	PVC
E_s (GPa)	3.0	2.3	2.9
E_d (GPa)	6.0	3.0	4.6
ρ (kg/m ³)	1180	1190	1430
ν	0.33	0.39	0.36

Table 2.2.1: Static and dynamic material properties at room temperature

2.2.2 Brief review of experimental methods on RCP in polymers

As already stated in the previous chapter, G_{ID} is related to the crack propagation phase and reaches its minimum value under plane strain conditions. Introduction of RCP into the brittle materials posed no problem, whilst the major obstacle in RCP experiments in tougher materials is to induce a brittle, unstable crack. Ductile and stable cracks were often generated. In some cases blunting of the crack tip was followed by gross yielding. Conventional loading techniques used for the actuation of RCP in tougher polymers are:

- a) Quasi-Static (QS) loading of a precooled sample,
- b) Impact loading of a single-edge notched bend (SENB) sample, and
- c) Tests on duplex samples, where the crack is initiated in a brittle starter section and propagated into a tougher region under QS loading.

An extended review of loading techniques in tough polymers is given by Genussov[89]. He developed a new 'glass-duplex' technique which overcomes the difficulties in monitoring and analysing impact tests. The technique incorporates a polished glass-slide bonded over the crack mouth of a single-edge notched, sidegrooved sample made of tough polymer. The sample is then loaded quasi-statically in tension until the glass breaks and a brittle crack initiates from the

rapidly loaded crack tip. This technique is used in the present work for dynamic fracture testing of PC and PVC, and the details about the method and specimen preparation are presented in section 4.2.2.

Most of the reported work on dynamic fracture considers brittle materials such as Araldite-B, Homalite and PMMA, and standard QS fracture tests have been used successfully to initiate RCP in these materials. There are three principal investigators in this field:

- a) Kalthoff[80, 83, 85] - employed wedge loaded single-edge notched (SEN), compact tension (CT) and double cantilever beam (DCB) specimens,
- b) Kobayashi[78-b, 84, 86] - studied QS loaded samples such as wedge-loaded DCB, fixed grip single-edge notched tensile (SENT) and CT in dynamic fracture investigations of various materials, and
- c) Ravi-Chandar and Knauss [84-b, 85] used a magnetic pulse method of step loading the crack faces of large SEN specimens.

A number of authors employed the SEN specimen for its simplicity (section 3.2). Some of them, e.g. Carlsson[73], used large specimens because the idealized analytical solution for an infinitely large plate might then be applied (section 1.2.2). Bergkvist[73] proposed the use of a specimen with high width to height ratio, since analytical solutions also indicate a negligible dynamic correction for the fracture of an infinitely long strip sample of finite height.

2.3 Methods of recording dynamic fracture behaviour

2.3.1 Direct toughness measurement

RCP properties can be characterized by employing either direct or indirect techniques. This section briefly discusses the principles of some of the direct methods. These involve measurements of local crack tip features, such as surface deformations, temperature due to a heat generation and CTOD, from which K_{ID} (G_{ID}) data are obtained. Most of the direct techniques utilize one of the idealized analytical solutions for the extraction of K_{ID} (G_{ID}) data from experimental records, i.e. dynamic stress intensity factors are computed at discrete points along the crack path by fitting a postulated static or dynamic crack tip field to the whole field experimental data. Indirect methods rely on the crack length history and boundary conditions, monitored during the test and used later, usually in numerical simulations, to generate G_{ID} or K_{ID} . The techniques of both classes require precise knowledge of the material properties. Having discussed the uncertainties and possible sources of errors involved in the generation of RCP properties by direct methods (section 1.4), the author favours the indirect, coupled experimental-numerical technique which is the one used throughout this work (chapters 5, 6).

Shadow optical caustics and fringe-pattern photoelasticity are the most frequently used direct, optical techniques. Both involve high speed photography of the sample during RCP and can be operated in a transmission mode of lighting (in the case of transparent materials) or in a reflection mode (for opaque materials).

Dynamic photoelasticity observes the planar surface deformations in a wider zone around the crack tip (Kobayashi[86], Dally et al [85], Green and Pratt [74], Knauss and Ravi-Chandar [85]). Using polarized light the crack tip stress field may be visualized as isochromatic shear

stress loops. The method for extracting K_{ID} data from the isochromatic data involves a calibration based on the static or dynamic solutions for the state of stress near the crack tip. This method, which can only be applied to a limited range of materials, may also be extended to steel samples by adhering a photoelastic (birefringent) surface coating (Kobayashi and Dally [80]).

Beinert and Kalthoff [81] described, in detail, the method of Caustics (also called the method of shadow patterns and the shadow spot method). This method is very widely used to analyse dynamic fracture events (Kalthoff[80, 83], Beinert and Kalthoff [81], Rosakis and Ravi-Chandar [86], Shockey et al [83]). The method utilizes the information from the stress strain field within a small area close to the crack tip. Knowledge of external loading and geometric boundary conditions is not necessary when this method (or dynamic photoelasticity) is used. The stress intensification in the region around the crack tip causes a reduction of both the thickness of the sample and the refractive index of the material. Consequently, the light transmitted through the specimen is deflected outward (the laterally deformed region around the crack tip acts as a concave lens). The light deflection in the reflection mode is caused only by the thickness change of the sample. In the shadow image of the crack, the crack tip appears surrounded by a dark spot bounded by a bright light - a caustic. A diameter of this round shadow pattern is related to K_{ID} through one of the idealized infinite plate solutions (section 1.2.1, 1.2.2).

Another direct method of recording RCP properties is the application of strain gauges near the crack path (Shukla et al [89], Dally and Sanford [89], Sanford et al [90], Sun[83]). Strain gauges are adhered along the prospective plane of crack advance and the strain histories and the crack speed are recorded during RCP (section 6.3.2). Then, each of the strain vs time readings is fitted by an idealized infinite plate prediction yielding K_{ID} data (at the discrete points corresponding to the strain gauges positions).

In materials like metals and tough polymers nearly all the energy absorbed by a moving crack is due to plastic deformation. This plastic energy is dissipated as heat and measurements of the temperature rise near the running crack can be used to calculate G_{ID} by solving the equation of heat flow. The temperature measurement can be achieved by using a row of thermocouples placed along the prospective crack path (Shockey et al [83]). A more rigorous technique is reported by Rosakis[89] who utilized high speed infrared sensors to measure the temperature rise inside the plastic zone of the moving crack.

K_{ID} may also be correlated to the crack tip opening displacement (CTOD) measured during the crack propagation (Kobayashi and Dally [80], Kanninen and Popelar [85], Suzuki[91]). The CTOD can be determined by employing various pieces of equipment such as high speed cameras, eddy current displacement transducers or optical fibers. Several correction factors for K_{ID} calibration from CTOD are reviewed by Kanninen and Popelar [85].

2.3.2 Crack speed measurement

As mentioned in the previous section, an indirect method of characterizing RCP parameters is used in this study. However, both the direct and indirect methods require an accurate crack length measurement for the appropriate interpretation of G_{ID} (K_{ID}) vs $\dot{a}$ and/or the precise generation of the RCP properties (Dally et al [85], Rosakis and Zehnder [85]).

Several methods, which can be divided into continuous and discrete (incremental), have generally been employed to monitor the crack speed. These include electrical resistance methods, ultrasonic fractography, fast photography and an application of optical fibres. A thorough review of the techniques of crack speed measurements is given by a number of authors, e.g. Erdogan[68], Rosakis and Zehnder [85], Genussov[89]. One of the major problems encountered in recording the crack history data is the identification of the exact crack

tip position. This can be due to several reasons such as the formation of microcracks or microvoids ahead of the main crack, the curved crack front (crack tunneling), and the large strains near the crack tip (strain-gauging can obscure the electrical methods readings). The incremental crack monitoring techniques have two main disadvantages:

- a) the difficulty in monitoring the crack behaviour right after the initiation - the duration between successive readings is often too long and the details of the crack initiation are usually missed,
- b) the generation of the continuous crack history data - if the interval of measurements is not small enough, errors may result due to variations in the crack speed on time scale smaller than the time interval.

Due to its accuracy the fast photography continuous technique is used in this study for recording of RCP in duplex PMMA/PC SENT samples. Later, as the crack speed was found to be constant in the tested materials, the electric grid line method is employed because of its simplicity. These two techniques are briefly described below.

A simple resistance grid-line technique has been used by many researchers in dynamic fracture of polymers. The method consists of a grid of conductive material of some type (usually graphite or silver paint) placed on the specimen surface along the prospective crack path and connected electrically in parallel. The conductive material must be mechanically compatible with the sample material - both should break simultaneously. As the crack advances, the conducting lines are broken (one by one) and the resistance of the whole grid is increased. A step voltage response, in the total resistance reading, is obtained when each line breaks. No calibration is needed since the exact positions of the lines are known and the step voltage response amplitude is of no significance. Only an average crack speed between lines can be measured. For the constant velocity crack propagation, as obtained in the present study, this is not a problem. More accurate crack speed measurement can be obtained by increasing the number of the

conducting lines. The number of the grid lines is limited since difficulties may be experienced in resolving each voltage step (the magnitude of each resistance step decreases with increasing number of the grid lines). This problem is overcome and the resolution of the resistance step readings is improved when the lines are connected in four groups as arms of a Wheatstone Bridge (Peretz and Dibeneditto [72]). More details of this technique are given in section 4.3.

Fast photography is the most widely used method of monitoring RCP events. The method is conducted by employing a high speed camera to capture the rapidly propagating crack history data. Two modes of fast photography technique can be operated: framing and streak photography. Framing mode gives discrete images of the sample during a fracture event whilst streak mode produces a continuous picture of the crack history. Near the interface between two different materials a 'discontinuous' crack behaviour might be expected. Thus, in this work, streak mode photography is used to record RCP in duplex PMMA/PC specimens. The details of the experimental setup and high speed camera are presented in section 3.4.

Both grid-line and fast photography techniques require a triggering device to start the recording of the crack length vs time data (sections 3.4, 4.3).

2.4 Review of some experiments related to RCP in multi-layer systems

This section provides some insights into the problem of the rapid cracking of multi-layer systems. Experimental studies on the embrittlement of ductile materials are reviewed in the first part of this section. The concluding part considers reported works on duplex specimen geometry.

Typical behaviour of ductile materials, fracturing at room temperature (RT) under slow loading rate conditions, is generally associated with large applied strains, stable slow crack growth and substantial plastic deformations (figure 2.4.1).



Figure 2.4.1: Typical tensile load-displacement curve for ductile fracture.

By lowering the test temperature and/or by increasing the rate of loading, normally ductile polymers may fail in a brittle manner (figure 2.4.2).

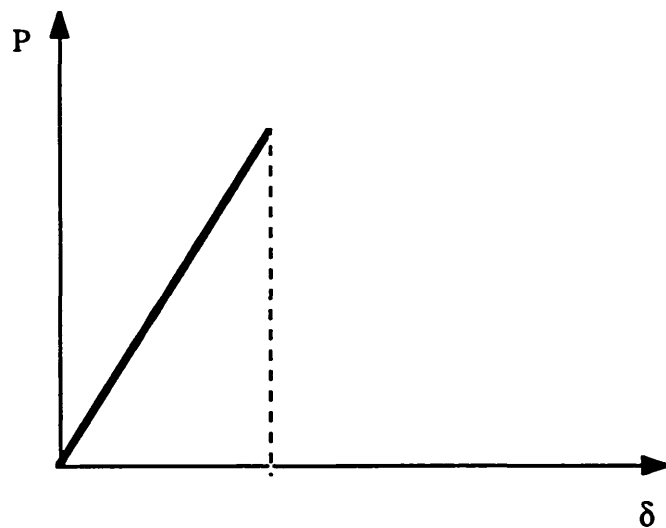


Figure 2.4.2: Typical tensile load-displacement curve for brittle fracture.

A brittle material, used as a part of a multi-layer system, could promote an unstable, brittle fracture of the more ductile part of the system, even when RT and slow loading conditions are retained throughout the test. That brittle/ductile material system can arise as a result of aging, exposure of a ductile material to ultraviolet (UV) radiation or chemical reaction, an application of a brittle coating on a ductile material, a bonding of a brittle material onto a surface of a more ductile one, etc.

2.4.1 Embrittlement of ductile materials

The mechanical behaviour of polymeric materials can be greatly influenced by the nature and properties of their surfaces. The studies of the surface embrittlement are related to the problems wherein the ductility of polymers is drastically reduced by the presence of the brittle layer. A 'thin' brittle layer at the surface of a ductile polymer can cause failure of the bulk material at much reduced applied strains.

It was shown that the thickness of the brittle layer had an important influence on the surface embrittlement of the ductile part of the system. Also, the toughness of the brittle material introduced on the surface of the more ductile material, was found to play a significant role.

Priebe and Stabenov [74] measured the Charpy impact strength of acrylonitrile-butadiene-styrene (ABS) coated with styrene-acrylonitrile (SAN). They concluded that there was a critical thickness, in the order of 10-20 μm , at which loss of the ABS toughness occurred.

Bucknall and Street [68] reproduced the effect of the outdoor aging on rubber modified thermoplastics (the ABS and high impact polystyrene - HIPS) by laminating a layer of a glassy polymer onto the surface of unaged samples. They showed that the aging reduced the energy of the crack initiation and the impact strength was determined by the crack propagation energy.

Bucknall and Street [68] suggested that the surface layer was equivalent to the introduction of a sharp notch.

Djiauw and Fesko [76], and So and Broutman [82] performed a similar experimental analysis to show the effect of brittle coating thickness on the tensile properties of some ductile materials. Djiauw's work shows the effect of the temperature, loading rate and paint film thickness on the behaviour of various automotive-grade substrates (an olefinic-base thermoplastic elastomer TPN Nordel 3745, a urethane-base thermoplastic elastomer TPU Estane 58130). The paint coatings were considerably more brittle than the substrates and they displayed relatively low ultimate elongation and substantially higher stiffness. Depending on the rate of loading, test temperature and paint film thickness, stress-strain curves for the various paint-substrate systems exhibited either ductile or brittle fracture behaviour in high speed tensile tests. The substrates alone always exhibited ductile behaviour under the same test conditions. The ultimate elongation of the TPU substrates was sharply reduced to the value corresponding to the ultimate elongation of the coating alone. Although a reduction of the ultimate strain with increasing coating thickness was found in coated TPN samples, these failed in a ductile manner and displayed multiple cracks on the coated surface. For thin films, the coating did not severely perturb the crack growth in the elastomers although it provided an initial crack. For thick coatings, the higher modulus coating stored considerable energy prior to failure. When the sudden crack initiation was combined with high energy stored in the substrate, the crack propagated catastrophically and the elongation of the painted TPU became that of the paint alone. The TPN is different in that it undergoes elongation at a much lower stress level (low yield stress) so that insufficient energy was available at the time the coating failed to propagate the crack through the specimen. Subsequently, other areas of the coating reached the ultimate elongation and a network of cracks was formed before one of them propagated entirely through the sample. Since the coating was broken in several places, the stress concentration was reduced in the individual cracks so that the specimen achieved higher strains.

So and Broutman [82] performed an experimental study in order to clarify the mechanism of the surface embrittlement of rubber-modified thermoplastics. Brittle polymer films were laminated onto the surface of ABS and HIPS samples to simulate either embrittlement due to the environmental degradation or brittle paint application. It was shown that the breaking strain of the ABS and HIPS continued to decrease as the coating thickness increased up to a certain value, called the critical thickness. Thereafter, the breaking strain was reported to remain independent of the coating thickness and equal to the failure strain of the coating. Thus, at the critical coating thickness, the sample failed in a brittle manner as evidenced by the absence of an apparent yield point. They ([82]) concluded that the critical coating thickness is equal to the critical crack dimension just prior to the onset of unstable crack growth. When the coating thickness was less than the critical crack size required for unstable crack propagation, only slow, stable crack growth could take place.

The results of experimental studies performed to evaluate the toughness of some plastic coating materials were reported by Moore and Prediger [86]. They assessed the thickness of epoxy coating applied on a PC substrate which would ensure ductile behaviour of epoxy/PC laminates under impact loading. It was assumed that brittle cracking can occur if the plastic zone developed in the coating is an order of magnitude smaller than the coating thickness. The assumption was supported by comparing coating thickness, which corresponded to the brittle fracture of laminates under impact tests, with the statically determined plastic zone size at the tip of the crack.

Floyd[83] analysed the cracking of the latex paints applied to exterior wood substrates. Due to differences in the coefficients of expansion along the wood grains, cracks were generated and propagated in the wood during exposure to weather cycles. Inclusions in the paint film stretched across the wood cracks acted as stress concentrators, leading to the initiation of paint cracks. A two parameter model was introduced, employing the fracture energy of the paints and the liquid water permeability, to account for the cracking behaviour of the composite. It was shown that a composite with a low fracture energy required a low strain energy and,

therefore, propagated cracks more easily.

Broutman[81, 82, 83] pointed out that a brittle surface layer (induced either by oxidation, UV radiation or during processing) could shorten the expected service life of a tough and ductile pipe material - polyethylene (PE). He claimed that when the affected layer was thin, the crack arrested when it reached the unaffected material. As the deteriorated surface layer continued to grow thicker, a crack which initiated in the brittle layer, propagated into the PE material. With continued application of the load the crack sustained its advance through the PE pipe wall.

The influence of paints on the mechanical properties of coated plastics was investigated by Maier and Laible [88]. They found that an unsuitable coating system drastically reduced the impact resistance of the painted part. The dimensional stability of the coated plastic part was negatively affected by the difference in the coefficients of expansion of the coating layer and the plastic substrate. By using a two-layer primer system, which allowed a graduated coating flexibility from highly flexible primer over a semi flexible primer surfacer to the hard topcoat, the impact resistance of the coated part was reported to retain a high level.

Most of the experimental studies reviewed in this section employed unnotched samples made out of a ductile material with a thin layer of brittle material introduced on its surface. The loss of the toughness of the ductile material, due to the presence of the surface brittle layer, was assessed by using either tensile or impact loading techniques. Various fracture mechanisms were proposed in order to explain the surface embrittlement of different ductile materials. The critical thickness was defined as the thickness of the brittle layer at, and above which, a brittle/ductile system fails in a brittle manner at a breaking strain equal to the ultimate strain of the brittle material alone. The main conclusion was that due to the introduction of a brittle layer onto the ductile material surface, the crack initiation energy of the ductile material was reduced and its strength was determined by the propagation energy of a brittle crack.

2.4.2 Reported work on duplex specimens

A duplex specimen (DS), consisting of a brittle material laminated onto a surface of a more ductile material, has generally been employed for two purposes; to simulate the surface embrittlement of ductile materials and to introduce RCP into ductile materials using a brittle layer as a RCP starter section. There are two main variations of the DS geometry:

- a) a prenotched brittle phase is attached to an unnotched ductile phase,
- b) an unnotched brittle layer is adhered to a prenotched ductile material.

In the present work, the analysis of RCP in the PMMA/PC and the PMMA/PVC two-layer systems is performed by using the DS configuration-(a), while the DS geometry-(b) is employed to initiate a fast crack in PC and PVC.

This section briefly discusses relevant experimental work on the duplex specimen.

Rolland[83], and Rolland and Broutman [85] utilized a brittle material/ductile material DS in order to model the surface embrittlement arose from the application of either PMMA or styrene acrylonitrile (SAN) onto surfaces of PC or acrylonitrile butadiene styrene (ABS) samples. An unstable crack was initiated in a prenotched brittle part of a SEN duplex sample under slow tensile loading. The speed of the crack approaching the interface between the two materials was measured by using a conductive paint grid applied on the brittle part of the DS near the interface. They claimed that there was a critical crack velocity above which the crack passed through the interface, and below which it arrested. For example, for the PMMA/PC DS with a thickness of 9.8 mm the crack was reported to propagate across the interface into the PC only when the crack speed was above 320 m/s. The critical crack velocity, which corresponded to the critical thickness and toughness of the brittle material, was also reported by So and Broutman [86], as a condition for surface embrittlement. As the surface layer toughness increased, they found that the crack speed requirement decreased. Tougher ductile materials

needed a greater critical crack speed. When the crack reaches the brittle-ductile interface, it can either arrest at the interface, propagate across and arrest in the ductile material, or propagate completely through the sample. It was predicted that in order for the crack to propagate across the interface it was necessary that:

$$[(K_{ID})_{\text{brittle}}]_{\text{interface}} > (K_{IA})_{\text{ductile}} \quad 2.4.2.1,$$

where, K_{IA} , the crack arrest energy, is the minimum value of K_{ID} of the ductile material. The above assumes a good interfacial bonding, i.e. no delamination as the crack propagates across the interface. The condition 2.4.2.1 is schematically presented in figure 2.4.2.1.

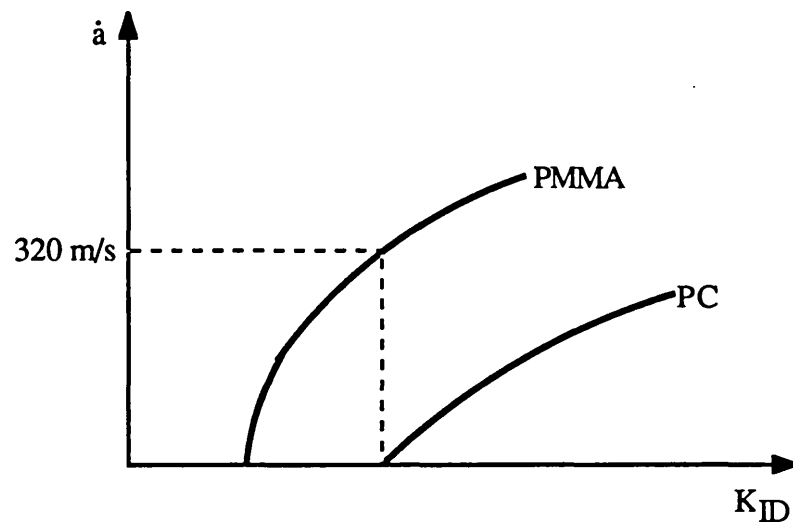


Figure 2.4.2.1: Schematic presentation of the condition for surface embrittlement
(after Rolland[83])

Rolland[83, 85] obtained dynamic fracture toughness data of the PMMA and PC using a wedge loaded compact DCB specimen. The calculation was entirely based on the static LEFM prediction in the form of equation 1.2.1.3 given by Crosley and Ripling [80], with no dynamic correction included. From K_{ID} vs $\dot{a}$ curves of the PMMA and PC (figure 2.4.2.1), Rolland concluded that for criterion 2.4.2.1 to be fulfilled, the velocity of the crack reaching the PMMA/PC interface had to be around 320m/s. This crack speed was equal to the critical speed

measured from the PMMA/PC DS (Rolland[83, 85]).

This is a most surprising result since the accuracy of K_{ID} data and the validity of the prediction 2.4.2.1 are very doubtful. There are quite a few uncertainties concerning the fracture toughness data reported by Rolland[83, 85]:

a) The shear strain (γ) along the neutral axes of the two arms of the specimen was assumed to be independent of the crack length (a) and dependent only on the applied load (P). The static calibration curves γ vs P and C vs a (C is the specimen compliance) were used to determine the crack opening displacement (Δ) at the edge of the sample during RCP, since $\Delta=CP$. Then, the crack length, recorded by employing a conductive paint grid technique, and the corresponding Δ were used in the static prediction to generate dynamic fracture toughness at different moving crack tip positions.

Sections 1.2.2 and 1.4 discuss the difficulties in dynamic fracture calculations, related to the effects of the inertia and of the stress waves propagation. It is demonstrated that, due to these dynamic effects, the dynamic stress intensity factor and the dynamic energy release rate cannot be accurately derived from the crack length, loading and geometry conditions as in static fracture events. Kanninen[85] showed that the static approach to the crack propagation and arrest events generated significant errors, especially when RCP in a DCB specimen was analysed. The comparison of the statically and dynamically computed stress intensity factors revealed the considerable difference (up to 300%) for both propagating and arresting stages of the crack in a duplex DCB specimen.

b) K_{ID} vs $\dot{a}$ curves of the PMMA and PC reported by Rolland[83, 85] are not in agreement with reported fracture toughness results (Crouch[86], Rosenfield and Kanninen [73], Kobayashi et al [80]), obtained by a more sophisticated, either direct or indirect, dynamic analysis. For example, the PC dynamic fracture toughness obtained by Rolland is approximately 30% less at $\dot{a} = 0$ m/s and 50% higher at $\dot{a} = 410$ m/s than K_{ID} of the PC presented by Kobayashi et al [80]. Thus, even the shape of the K_{ID} vs $\dot{a}$ curve generated by

Rolland is different than usually reported.

c) Both the PMMA and PC are viscoelastic materials. γ vs P and C vs a calibration curves, used to generate dynamic fracture toughness data, were obtained under slow loading while RCP in the PMMA and PC occurred under very high rate of unloading (section 5.2.4).

For a propagating crack, the fracture criterion is defined by equation 1.2.2.6 (section 1.2.2). As long as the crack propagates, this condition is satisfied at any crack position. When the crack, initiated in the PMMA part of the DS, reaches the PMMA/PC interface, it will propagate across if the condition 1.2.2.6 is fulfilled at the interface, i.e. if:

$$(G^{dyn})_{interface} = (G_{ID})_{PC}$$

$$(or\ equivalently\ (K^{dyn})_{interface} = (K_{ID})_{PC}) \quad 2.4.2.2.$$

According to the dynamic fracture approach to RCP (section 1.2.2) the crack will arrest at the PMMA/PC interface if the inequality 1.2.2.7 is satisfied, i.e.:

$$(G^{dyn})_{interface} < (G_{IA})_{PC}$$

$$(or\ equivalently\ (K^{dyn})_{interface} < (K_{IA})_{PC}) \quad 2.4.2.3.$$

Conditions 2.4.2.2 and 2.4.2.3 apply only when an adequate interfacial bonding is achieved. This was already acknowledged by Crosley and Ripling [73], who studied the crack arrest. They employed A533B steel specimens provided with a brittle weld. This enabled crack initiation in the weld at a value of K_{IC} below that of the base material. Thus, once initiated, the crack propagated through the brittle weld until it reached the interface. Crosley and Ripling [73] measured the crack arrest toughness of the steel by using a static approach, and they concluded that at the interface point, the crack arrested when K^{dyn} was less than K_{IA} of the base plate. On the contrary, the crack continued to propagate through the base material when K^{dyn} was greater than K_{IA} of the base plate.

Regarding generally accepted dynamic fracture criteria (equations 1.2.2.6, 1.2.2.7), it is not clear how and why the PMMA fracture toughness (property of PMMA) can become a condition for the crack propagation in the PC, as predicted by Rolland [83, 85] (equation 2.4.2.1 and figure 2.4.2.1).

Rolland[83, 85] pointed out that in order to embrittle a ductile material, a brittle material should not have a fracture toughness which is significantly lower than that of the ductile part, since the crack would not have enough energy to propagate through the interface. On the other hand, when K_{ID} of the brittle layer is too high, the crack might have difficulties initiating and propagating. It was speculated (Rolland[83, 85], So and Broutman [86]) that the brittle layer must be sufficiently thick for the crack to grow beyond the critical dimension, and so, to accelerate to the critical crack speed when it reaches the interface (the critical crack dimension was defined as the length of the crack at the onset of unstable growth).

The DS consisting of a brittle starter material attached to a more ductile one, has also been used in order to initiate a fast crack in the ductile part of the sample.

Hoagland et al [77] successfully used the DS configuration produced by welding a hardened steel starter section to the A533B steel test section. The aim of his work was the determination of the crack arrest properties of the test section. Dally and Kobayashi [78] employed dynamic photoelasticity to study crack arrest and reinitiation in duplex specimens. The duplex samples consisted of Homalite 100 as the starter section and a tough epoxy as the arrest section. The two phases were cemented together with a tough, high shear-strength adhesive. The main observation was that the crack abruptly arrested as it reached the adhesive layer between the two materials. During the crack arrest the stress intensity factor (K^{dyn}) increased rapidly as the kinetic energy in the specimen was converted to strain energy. If K^{dyn} reached the toughness of the adhesive layer, the crack reinitiated in the adhesive and penetrated into the arrest section of the duplex sample. As the crack initiated in the second phase the stress intensity factor

decreased. The crack propagated at a higher speed in the starter than in the arrest section. If K^{dyn} did not achieve the required value, the crack remained stationary while K^{dyn} oscillated. Crouch[86] unsuccessfully employed a duplex SENT sample of brittle and tough grades of the PE fused together. The crack arrested shortly after penetrating the tough PE grade, probably because of inadequate adhesion between the starter and test section. Furthermore, trying to inject an unstable crack into the PE, he used duplex PMMA/PE TPB samples. This setup failed to generate RCP in the PE since the brittle fast crack in the PMMA arrested upon entering the PE. Having no external driving force to propagate the crack forward, the amount of released energy was insufficient for unstable crack jump into the PE.

Theocaris and Milios [80, 81] studied the crack propagation in two-phase epoxy resin specimens by the method of caustics. The samples were composed of two layers cast in two steps from the same epoxy resin, but containing different amounts of plasticizer. There was no adhesive layer between the two phases. The crack was reported to propagate at constant speeds in both phases. The crack speed in the second phase decreased with increasing ductility of the second phase. This speed was found to be dependent on the material characteristics of both phases. It was also influenced by the sudden crack arrest which always occurred at the interface. It was reported that K^{dyn} increases sharply as the crack approaches the interface and decreases during the period of crack arrest. After the crack reinitiation in the second phase, K^{dyn} rises again up to a new stationary value.

Genussov[89] suggested an inverse duplex SENT sample. Instead of adhering a prenotched brittle layer to an unnotched ductile PE, an unnotched brittle layer - glass slide, was fused over a prenotched PE sample. That way, the unnotched glass did not break easily under remote tensile loading (as happened to the PMMA/PE DS in Crouch's work), and a large amount of strain energy was stored in the sample. Since the brittle layer was very stiff compared to the PE, it carried most of the load off the crack tip which remained sharp. Consequently, sudden breaking of the glass resulted in brittle crack initiation and propagation.

CHAPTER 3

EXPERIMENT ON DUPLEX PMMA/PC SENT SPECIMEN

3.1 Introduction

The first two chapters provided some insight into RCP related problems and discussed the difficulties associated with the characterization of dynamic fracture properties. Acknowledging the observations reviewed in the previous chapters, an indirect technique of analysing moving crack behaviour is employed throughout the present study. This is a combined experimental-numerical G_{II} calibration method which relies on experimentally recorded crack length history and loading and boundary conditions.

The experimental work on the duplex PMMA/PC SENT sample is considered in this chapter. Firstly, the geometry of the DS and its preparation are described in section 3.2, followed by the discussion of the results of preliminary tests on the PMMA and PC single and duplex specimens (section 3.3). Section 3.4 presents the details of the crack speed measurements in the PMMA/PC DS. Finally, the DS crack speed results are summarized in section 3.5.

3.2 Duplex specimen details

The accuracy of G_{II} data generated by the indirect technique relies on the accuracy of the crack tip position and loading and boundary conditions recorded during RCP. The crack should be made to advance with a straight front producing a uniform, planar surface. This enables the exact crack tip location to be measured with a surface technique. It also enables the modelling of RCP by a lower cost two-dimensional (2-D) model. Again, in order to simplify and make the modelling of RCP more accurate, one should be able to determine precisely the boundary conditions of loading and deflection. Oscillations of the sample and its slippage on supports, friction of the wedge, bouncing of pins etc. should be suppressed during the crack propagation.

An SEN specimen, clamped rigidly between two heavy grips and loaded slowly in tension until RCP occurs, seems to satisfy these demands. Dahlberg et al [80] proved experimentally that heavy clamps, used to load the sample slowly until fracture occurred, ensured 'fixed grip' conditions during crack propagation (see section 6.2.2). During RCP in the fixed-grip SENT sample, the deflections are only due to the crack opening and there are no gross oscillations of the sample. Numerical modelling of this boundary condition is much easier than the modelling of boundary conditions of the impacted SEN bending (SENB) sample, the 'bouncing' pin loaded DCB or the wedge-friction loaded DCB. For example, numerical modelling of RCP in impact loaded SENB specimen should consider the dynamic effects related to the sample, the striker, the anvils and the contact stiffnesses between them.

Throughout the experimental part of the present work, the duplex SENT specimen (figure 3.2.1) was used for the testing of RCP in the two-layer PMMA/PC and PMMA/PVC systems. RCP in the PMMA was studied by testing SENT samples (section 4.2.1), while the glass-duplex SENT specimens were employed for the analysis of PC and PVC (section 4.2.2).

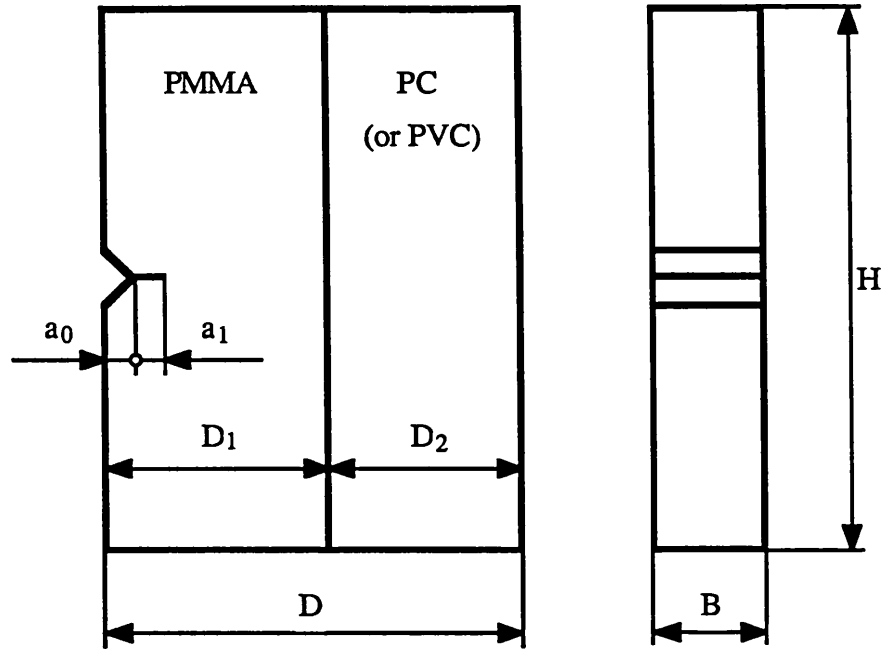


Figure 3.2.1: Duplex SENT specimen configuration

The steps involved in the preparation of the PMMA/PC duplex specimen are described below:

- a) The PMMA and the PC parts of the DS were cut from the sheets to the specified dimensions - each of the two parts of the DS required the same $B \cdot H$ (figure 3.2.1), while D_1 and D_2 were determined by the thicknesses of the PMMA and the PC sheets respectively.
- b) Before being adhered together, the surfaces of both the PMMA and the PC parts of the duplex sample were treated first by a sand-blasting technique and then with alcohol - ethyl alcohol (C_2H_5OH), 99.7% V/Vmin.
- c) A very good adherence between the PMMA and the PC was achieved by using the two component 'Tensol' Cement No. 70 (product of ICI); component A - methyl methacrylate mixture inhibited, component B - dimethyl phthalate. Component A was thoroughly mixed with component B, in the proportion of one gram of component A to one drop of component B. The mixture was left for approximately five minutes for trapped air bubbles to be released.

d) The pretreated surface of the PMMA part was 'painted' with a thin layer of the adhesive and was then bonded to the pretreated surface of the PC part. A pressure of about 0.025 MPa, was applied on each DS for at least 24 hours in order to obtain a good bonding as well as to get rid of remaining air bubbles.

e) The sides of the DS were polished to ensure uniform thickness (B) across the sample. Each sample was prenotched by milling a 90° 'V' notch of depth a_0 in the PMMA part of the DS. In a number of samples, a sharp notch of length a_1 was then introduced by pushing a sharp razor blade through the tip of the 'V' notch (figure 3.2.1).

Samples with varying precrack length ($a_0 + a_1 = 1$ to 3.5 mm) and bluntness were prepared and tested. This enabled an assessment to be made of the influence of the initial crack length and bluntness on the RCP in the DS. Also, duplex specimens of various dimensions (D_1 , D_2 , B) were tested in order to generate cracks of different speeds and to analyse the influence of the specimen size on G_{II} (section 1.4, Williams and Ivankovic [91], Ivankovic and Williams [91], chapter 5). Table 3.2.1 summarizes the average dimensions of the DS. The specimens were grouped in twelve sets, each of them containing approximately seventeen samples.

Set	D ₁	D ₂	B
TBW10	26.	9.4	10.
TBW6	"	5.8	"
TMW10	20.	9.4	"
TMW6	"	5.8	"
TSW10	13.	9.4	"
TSW6	"	5.8	"
TBN10	26.	9.4	6.9
TBN6	"	5.8	"
TMN10	20.	9.4	"
TMN6	"	5.8	"
TSN10	13.	9.4	"
TSN6	"	5.8	"

Table 3.2.1: The summary of the DS dimensions (dimensions in mm, H=100 mm).

3.3 Preliminary tests

A series of impact tests on both single and duplex PMMA and PC SENB specimens were performed prior to any study of RCP. These tests were conducted by W.Y.F. Chan [B.Sc. Final Year Project, Imperial college, 1989]. The main objectives of that work were to establish G_{IC} data of PMMA and PC, to check the quality of the duplex specimen preparation i.e. the quality of the PMMA/PC bond, and to assess the importance of the crack speed in the PMMA part with respect to the crack behaviour in the PC part of the PMMA/PC DS.

Single PMMA and PC SENB samples of various width - 13, 20, 26 mm and 5.8, 9.4 mm respectively, were tested. The thickness of both single and duplex SENB samples was 10 mm, while the height was 100 mm. Remaining dimensions of the DS are listed in table 3.2.1. The samples were prenotched by milling a 90° blunt 'V' notch of various depths (1 to 3 mm). A number of samples (including all the PC SENB) were provided with a sharp notch, 2 to 5 mm deep, made by pushing a new razor blade into the tip of the 'V' notch. One set of the tested duplex specimens had the initial sharp notch reaching the PMMA/PC interface.

The specimens were broken by using two types of impact machines; CEAST pendulum and ROSAND falling weight machine. The impact speed was set to 2 m/s. The filtering system, in both cases, was off to prevent the 'cut off' of the genuine signals.

Results from the impact tests have been analysed using a static solution, i.e. the energy release rate at initiation (G_{IC}) was calculated by employing equation 1.2.1.5. G_{IC} data of PMMA and PC were obtained as a function of the initial crack length. G_{IC} value of PMMA was found to be independent of the crack length and around $1 \times 10^3 \text{ J/m}^2$. The notch acuity showed a little effect on G_{IC} results of PMMA (brittle material). PC samples displayed a larger elongation at failure than PMMA samples, and calculated G_{IC} data of PC were around $4 \times 10^3 \text{ J/m}^2$. G_{IC} results of both PMMA and PC agreed well with reported G_{IC} data (Marshall et al [73], Williams[84]).

With the exception of the DS where the tip of the initial notch was at the interface, all other duplex specimens exhibited failures in a similar manner as the single PMMA samples. Load-displacement curves obtained from the tests on duplex specimens were similar to those of the PMMA samples. When debonding did not occur during the test, the elongation at failure of the PC was reduced to the elongation at failure of the PMMA alone (section 2.4.1). As expected, the DS with the tip of initial notch just 'touching' the PMMA/PC interface, failed in a similar manner as the single PC samples. The completely prenotched PMMA part of the DS could not

promote failure of the PC part at reduced elongation.

The conclusions from the preliminary test results can be summarized as follows:

- a) A quality of the duplex specimen preparation was found to be satisfactory (section 3.2). The PMMA/PC interfaces of broken samples were examined and debonding was observed in less than 5% of the tested duplex specimens. Debonding occurred in the samples where air bubbles were trapped at the interface.
- b) The failure of the PC part of the DS was considerably influenced by the crack behaviour in the PMMA part. A rapidly propagating crack, initiated in the PMMA part, promoted unstable fracture of the PC part at a reduced applied strain. Under the same loading conditions, the 'stationary' crack at the PMMA/PC interface did not affect the crack behaviour in the PC part. The importance of the crack speed in a brittle layer, with respect to the embrittlement of a ductile material, has already been discussed in section 2.4.2.

3.4 Setup and testing

The specimens were clamped rigidly between two stiff, heavy grips and loaded slowly in tension until RCP occurred. The weight of each grip was 14.35 kg while the stiffness was $1 \cdot 10^8$ N/m (section 6.2.2). The loaded boundaries were regarded as being under fixed grip conditions during RCP (sections 3.2, 6.2.2). Thus, global energy balance calculation of G_{II} (equations 1.2.2.2, 1.2.2.6) is based upon the initial loading conditions (prior unstable crack propagation) and the crack history data.

A conventional INSTRON tensile machine (model 1186) was used to break the specimens. The testing was performed under room temperature conditions (RT) and slow loading rate - cross head velocity (CHV) was 10 mm/min. The initial QS load and CHV were recorded by the internal Instron chart recorder. The measured cross head displacement was an overestimation

of the true sample displacement because of the wedge-type grips in which the samples were self tightened. However, this displacement reading was not essential in the simulation of RCP because the calculation of the available strain energy was based on the applied load and modulus (section 5.2.4).

The straight crack was obtained when the grips were free to rotate about their axes - no torsional load was then applied on the specimen.

The rest of this section describes the measurement of crack length histories by the fast photography technique. The details of monitoring the crack speed using the grid line technique, are presented in section 4.3.

The fast photography method necessitates a triggering device to start recording. For this purpose, the breaking of the conductive triggering line by the moving crack was used. Consequently, the details of the crack initiation event were not captured.

For the specimens containing only 'V' prenotches, three patterns of triggering lines were painted across the notch: a) two silver paint lines (Electrodag 915, high conductivity paint, product of Acheson Colloids Company, UK), b) one line at the center and c) a zigzag line (figure 3.4.1).

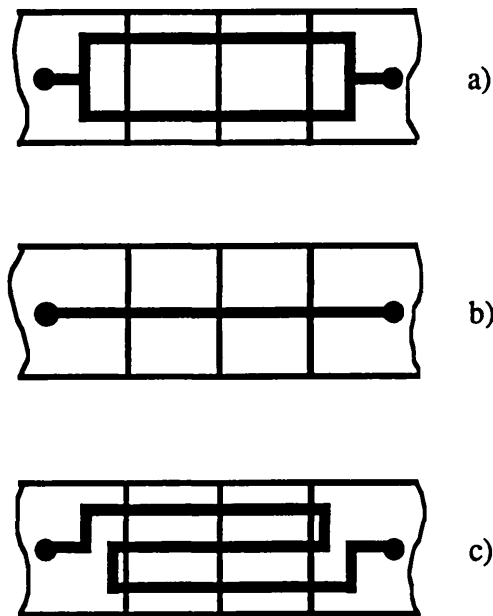


Figure 3.4.1: Patterns of triggering lines - 'V' prenotch

For the samples with sharp initial notches, silver paint was applied just in front of the tip of the sharp notch on: a) one side of the sample, and b) both sides of the specimen and connected in parallel (figure 3.4.2).

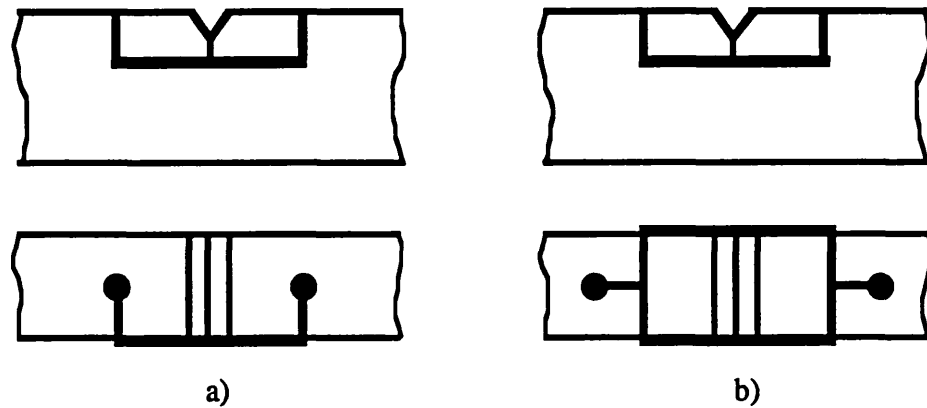


Figure 3.4.2: Patterns of triggering lines - 'V' + sharp prenotch

Both the PMMA and the PC are transparent materials. In order to obtain a clear, good contrast record of the crack length vs time, one has to ensure that the light from the flash passes only through the crack gap of the fracturing sample. For this purpose, a thin layer of opaque material has to be applied on the sample sides around the prospective crack path (figure 3.4.3).



Figure 3.4.3: Sample with black ink

This layer must satisfy several requirements:

- a) It must not affect the specimen surface and obstruct a 'real' crack propagation,
- b) The layer must fracture simultaneously with the specimen, i.e. the layer must be mechanically compatible with the base material.

Several kinds of brittle paints, graphite sprays and inks were used in order to meet these demands. Black ink seemed to give the best results and was used throughout this work.

The crack did not always start uniformly from the tip of the initial notch; it occasionally initiated from the sides of the tip. A possible reason is the non-uniform length of the prenotch through the specimen thickness. Obtaining a sharp prenotch of constant depth in the PMMA was extremely difficult since the crack often tended to propagate in front of the tip of the razor blade. Until both sides of the specimen were cracked, the flash light could not pass through the sample and the record was not clear. Thus, the best results were obtained with the two parallel lines triggering pattern (figures 3.4.1-a, 3.4.2-b) - a camera was triggered only when both lines were cut by the crack.

The crack length history was monitored by a high speed camera, model 790 IMACON, product of Hadland Photonics Ltd. The instrumentation was controlled by an adjustable, multiple delay generator (product of Hadland). This control unit was set on 'contact-break' mode, so that the breaking of the conducting line triggered two output signals. The first, with a time delay varying between 1-3 μs , triggered the flash and the second, with a time delay of 1-5 μs , triggered the high speed camera.

This camera incorporates a sinusoidally shuttered cathode tube which can operate in two modes: incremental - framing mode and continuous - streak mode.

In the framing mode, the electron beam carrying the image of each exposure is deflected to a different position on the screen. The rate of exposures is dictated by the plug-in units with 10, 25 and $100 \cdot 10^3$ frames a second. The smallest time distance between two adjacent frames is 10 μs . However, this time distance is far too large for recording of a discontinuous crack

propagation (section 3.5.2) in the DS and details of the crack behaviour near the interface were not captured.

Failing to monitor properly the crack length history in the DS with the fast photography technique operated in the framing mode, the streak mode was tried instead. In this mode, only a thin section of the sample is observed through a built-in slit in the camera. While the shutter is opened, the image is swept across the screen at a constant rate. The available streak plug-in unit has a variable sweep rate in the range of 1-10 $\mu\text{s}/\text{mm}$. In the present work, it was found that a sweep rate of 2-5 $\mu\text{s}/\text{mm}$ produced the best crack history results. The format of the streak picture is 70 mm horizontal (representing a time of 70-700 μs), and 18 mm vertical. The illuminated area illustrates the crack length $a(t)$ while the dark region represents the length of the sample ligament (figure 3.4.4).

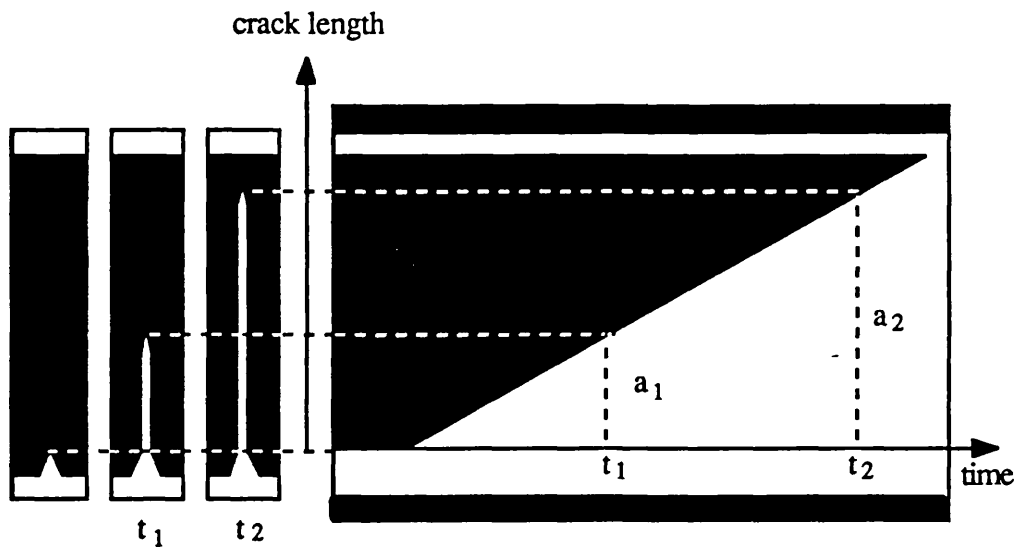


Figure 3.4.4: Streak image of a crack propagating at a constant speed

During the experiments, difficulties were experienced when the built-in slit in the camera was used. This was due to the vertical movement of the prenotch during loading, relative to the stationary camera. The camera had to be aligned below the crack in the unloaded sample for a

distance equal to the future crack path displacement during initial QS loading. This value was, of course, unknown and there was a high risk of losing the crack length history data. This problem was solved by cancelling the built-in slit in the camera. Instead, the sample was masked completely, exposing a very narrow section along the prospective crack path (figure 3.4.5).

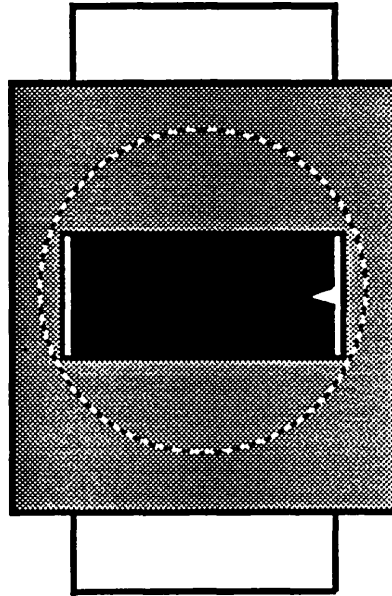


Figure 3.4.5: Masked sample

The sample was placed between the flash unit and the camera in a transmission mode (figure 3.4.6). In order to prevent any stray light from entering the camera a black tube was placed leading from the lens to the masked sample. Only the light coming through the crack in the sample reached the camera. As suggested by Genussov[89], a white paper diffuser was used between the flash and the sample, improving the contrast of the picture. Two gaps were left at the two sides of the sample for the light to come in - to match the sample's width, used as a reference length in the picture.

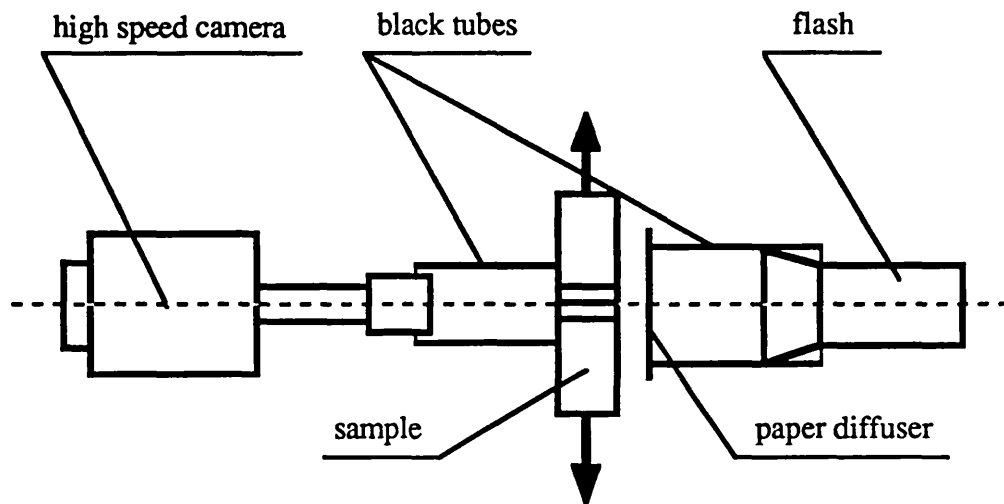


Figure 3.4.6: Photographic configuration

3.5 Crack speed in the duplex specimen

3.5.1 Note on the crack propagation behaviour

This section briefly reviews reported experimental observations on the crack propagation behaviour. Those observations are related to immediate post-initiation, bulk fracture and approaching arrest.

There are two main variations of reported crack behaviour immediately after initiation:

- a) the crack accelerates rapidly from zero speed to a maximum speed in only a few millimeters of the crack growth (Schardin[59], Crouch[86], etc.),
- b) the crack instantaneously accelerates to a maximum speed (Kobayashi and Dally [80], Genussov[89], Kobayashi and Ohtani [77], etc.).

The immediate post-initiation crack behaviour was found to be dependent on the bluntness of the initial notch. The crack was reported to accelerate instantaneously after completion of slow stable crack growth from a sharp notch, while from a blunt notch, the crack jumped instantly to a certain speed and then accelerated further at a measurable rate (Takahashi et al. [84]).

Most of the reported studies revealed remarkably constant crack speed over a significant portion of the specimen (Bradley and Kobayashi [71], Kobayashi et al [74], Crouch[86], Genussov[89], Wheel and Leever [91], etc.). The crack propagation at an approximately constant speed was observed in pre-loaded tensile tests where an impact initiated the crack (Paxson and Lucas [73], Genussov[89]), or even under reducing K conditions (Sun[83], Ravi-Chandar and Knauss [84-b, 85]). An interesting dynamic fracture test on the PE pipe with a variable thickness of the pipe wall was performed by Yayla[91]. He reported a constant RCP along the pipe sections of a constant wall thickness, while a sudden reduction of the crack speed corresponded to the abrupt reduction of the pipe wall thickness.

However, a limited number of authors reported that, during the propagation, the crack accelerated and/or decelerated (Takahashi and Arakawa [87, 90], Kalthoff[88], Rosakis and Zender [85]).

The experimental observations concerning the crack approaching the arrest can be summarized as follows:

- a) the crack gradually decelerates until arrest occurs (Kalthoff et al [76, 77], Takahashi et al [84], Bradley and Kobayashi [71], etc.),
- b) the crack arrest is apparently instantaneous (Bradley and Kobayashi [71], Kobayashi et al [76], Ravi-Chandar and Knauss [84-a], etc).

It is generally agreed that stress wave reflections from free lateral boundaries are relatively unimportant with respect to the crack behaviour. However, stress waves reflected from loaded boundaries could provoke crack speed fluctuations or arrest as they return to the crack tip (Rosakis and Zehnder [85], Dahlberg et al [80], Bradley and Kobayashi [71], etc.).

3.5.2 Duplex specimen crack speed results

A total of about 200 duplex specimens were broken. About 60% of the tested PMMA/PC DS produced satisfactory records of the crack length history. 40% of the specimens failed to generate the required results, i.e. there was no coherent record of RCP in the DS. These faults were mainly due to the improper setting of the recording instrumentation controlling the delay times for the flash and camera triggering, the sweep rate (speed) of the camera, the intensity of the light, etc. Furthermore, a good quality photograph of RCP was recorded only when the sample failure was associated with the smooth, planar fracture surface. However, enough high speed records were obtained in order to demonstrate the nature of RCP in the PMMA/PC system.

Typical streak mode photographs of crack length histories in the duplex PMMA/PC SENT samples are presented in figures 3.5.2.1-4. Figures 3.5.2.1-2 show RCP records of completely fractured samples - the cases where the crack propagated completely through both the PMMA and PC parts of the sample. The records presented in figures 3.5.2.3-4 demonstrate the histories of arrested cracks.

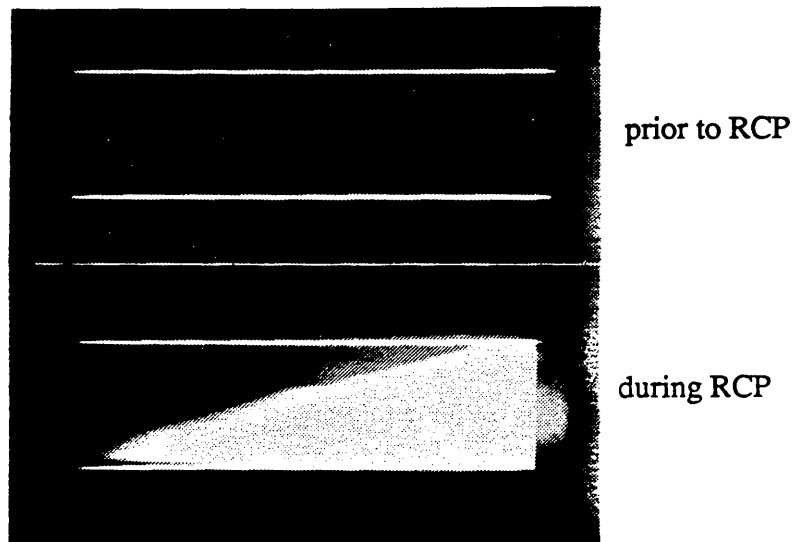


Figure 3.5.2.1: The streak mode photograph of completely fractured sample (TMN10-2)

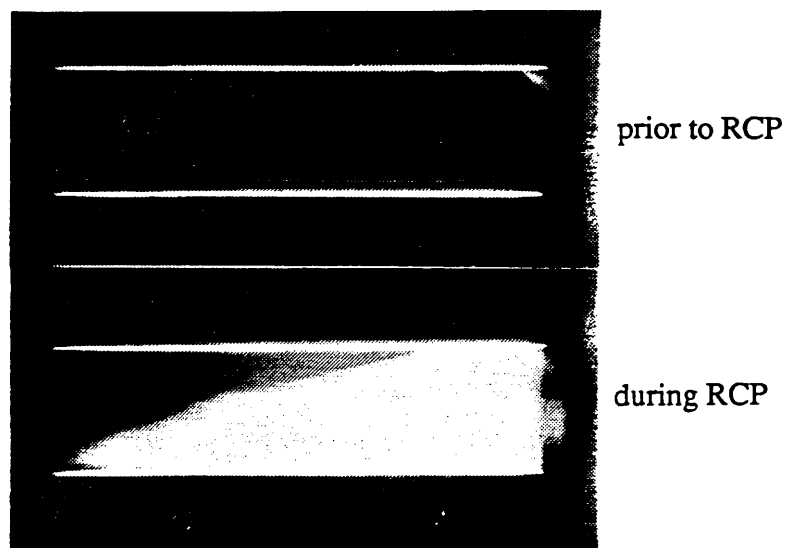


Figure 3.5.2.2: The streak mode photograph of completely fractured sample

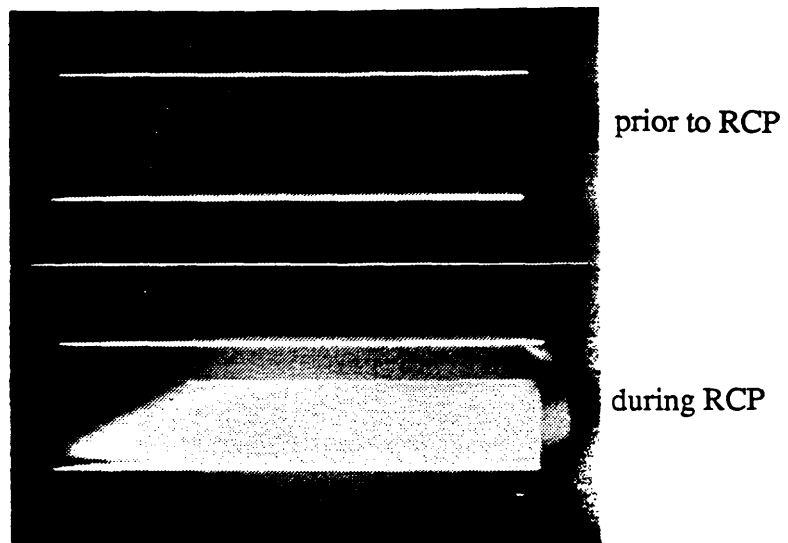


Figure 3.5.2.3: The streak mode photograph of a sample with the crack arrest at the interface

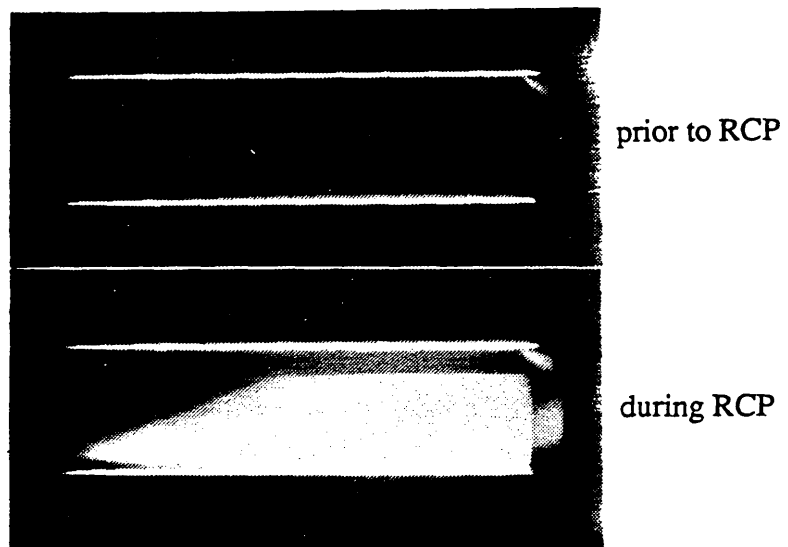


Figure 3.5.2.4: The streak mode photograph of a sample with the crack arrest at the interface
(TBW6-9)

Some important observations regarding RCP results of the duplex PMMA/PC samples are summarized bellow:

a) In each test a remarkably constant crack velocity was observed (figures 3.5.2.1-4). The recorded crack speed was in the range of 260 - 505 m/s in the PMMA, and 175 - 425 m/s in the PC part. RCP in the DS was characterized by higher crack speed in the PMMA part than in the PC part (figures 3.5.2.1-2, table 3.5.2.1).

b) Table 3.5.2.1 indicates that a higher average crack velocity in the DS corresponds to a higher breaking load (P_b).

test	$a_0 + a_1$ (mm)	P_b (KN)	$\dot{a}_{PMMA}$ (m/s)	$\dot{a}_{PC}$ (m/s)	comment
TBN10-2	2. + 0.	2.52	470.	230.	
TBN10-6	2.1 +0.	2.45	505.	175.	arrest in PC
TBW6-4	2.1 + 1.5	2.86	425		arrest at int.
TBW6-9	2. + 1.2	2.7	410.		arrest at int.
TMN10-1	2.1 + 0.	2.22	370.	305.	
TMN10-2	2. + 0.	2.06	344.	265.	
TMN6-3	2.1 + 0.	1.98	345.	233.	
TMN6-5	2.1 + 1.2	1.52	260.		arrest at int.
TSN10-1	2.1 + 0.	1.58	350.	240.	
TSN10-5	1. + 1.1	1.85	410.	313.	

Table 3.5.2.1: Selected results of the test on duplex PMMA/PC SENT samples

(int. stands for the PMMA/PC interface)

c) The crack never arrested in the PMMA while several crack arrests were observed in the PC. They are classified according to the duration of the crack arrest:

- The crack instantly arrested at the PMMA/PC interface for a very short time (3.4 - 9 μ s) and suddenly reinitiated in the PC at a high speed (figures 3.5.2.1-2, 3.5.2.5). This type of crack behaviour was common for all tests in which RCP occurred in both parts of the DS. The possible explanation is that the crack 'waited' at the interface until the condition for its rapid propagation into the PC (equation 2.4.2.2, section 2.4.2) was satisfied.

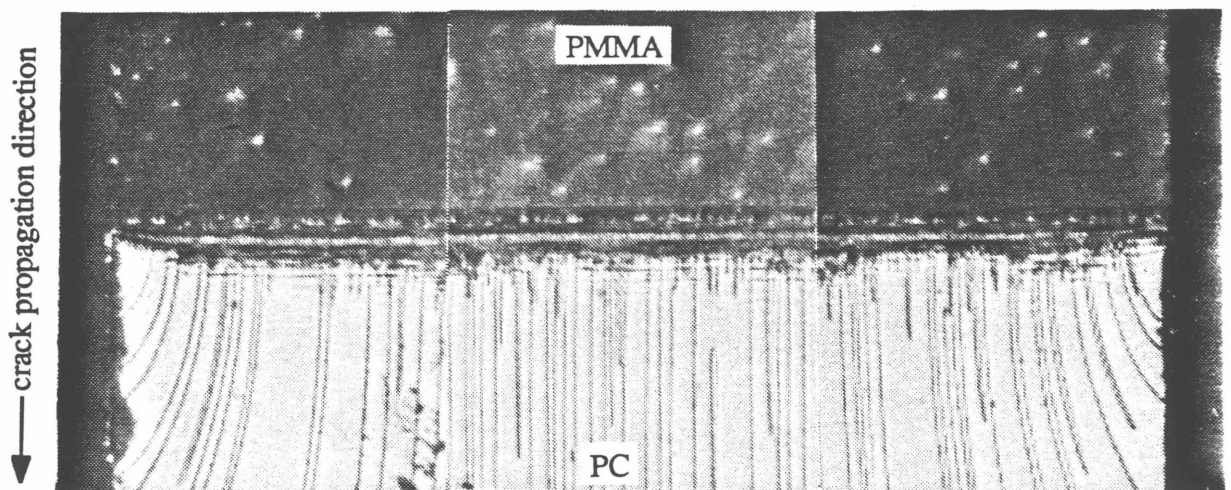


Figure 3.5.2.5: The crack arrest at the PMMA/PC interface for a short period

- Crack instantly arrested at the interface for a long time (figures 3.5.2.3-4, 3.5.2.6-7). It seems reasonable to speculate that the condition 2.4.2.2 (section 2.4.2) was not fulfilled. The crack behaviour was probably governed by the condition 2.4.2.3 and therefore, an unstable crack propagation did not occur in the PC. Eventually, as the applied load increased again, the stable crack growth took place in the PC (figure 3.5.2.7).

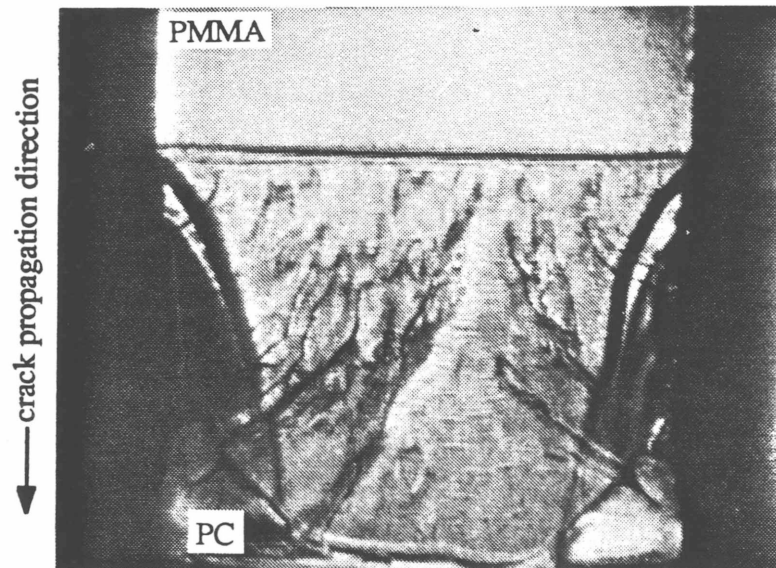


Figure 3.5.2.6: The crack arrest at the PMMA/PC interface for a long period

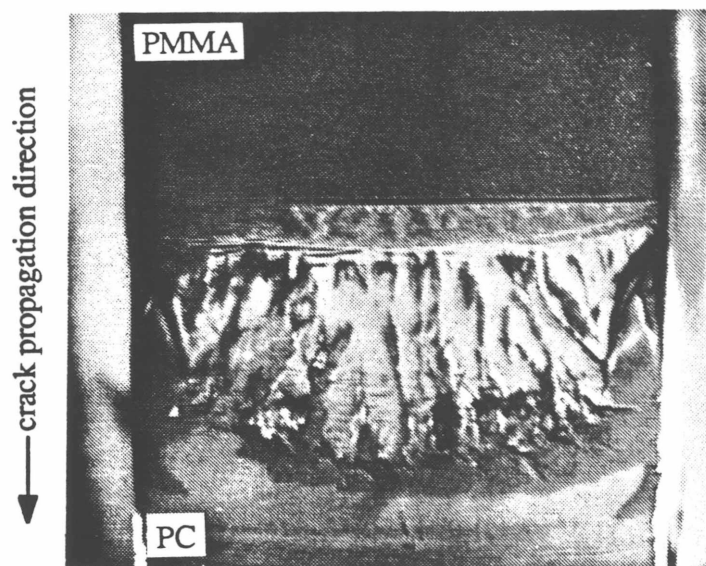


Figure 3.5.2.7: The crack arrest at the PMMA/PC interface for a long period

- The crack instantly arrested in the PC and reinitiated at the same speed as prior to arrest (figure 3.5.2.8). A remarkably straight crack front was associated with this type of crack arrest. According to the reported propagating crack behaviour (section 3.5.1), one can suggest that this discontinuous crack advance was provoked by compressive stress waves reflected from the loaded boundaries, as they returned to the moving crack tip.

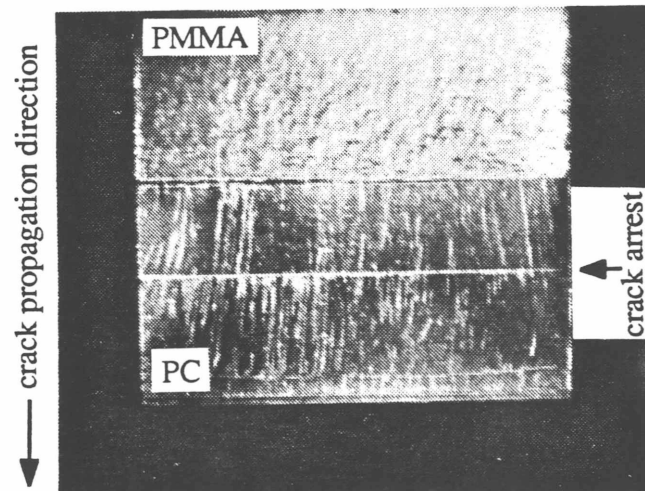


Figure 3.5.2.8: Instantaneous crack arrest in the PC part of the duplex sample

- The crack crossed the PMMA/PC interface, travelled some distance in an unstable manner and permanently arrested in the PC part of the sample. The crack front at arrest was curved near the sides and straight across the major portion of the specimen - crack tunneling (figure 3.5.2.9). The arrest originated near the specimen sides while the crack still propagated through the major section of the sample where plane strain conditions prevailed. A high speed record showed no evidence of crack deceleration prior to arrest. The recorded crack length at arrest is an underestimation of the true crack length because of the curved shape of the crack front (figure 3.5.2.10).

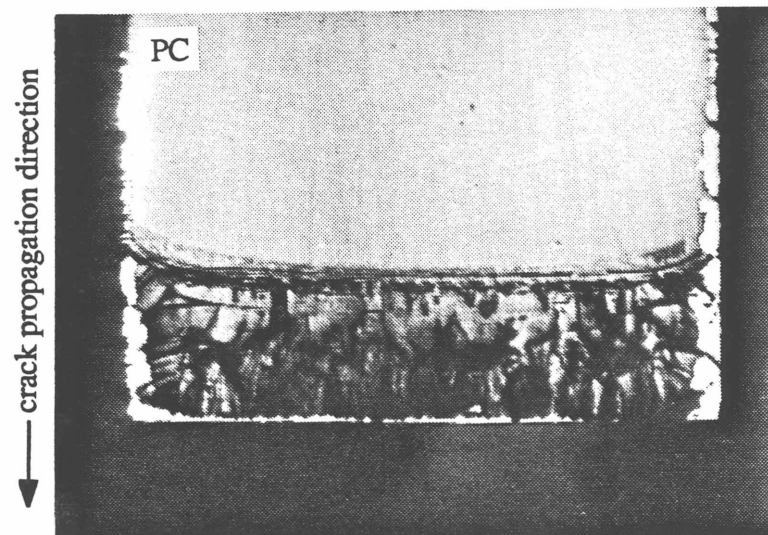


Figure 3.5.2.9: Permanent crack arrest in the PC part of the duplex sample

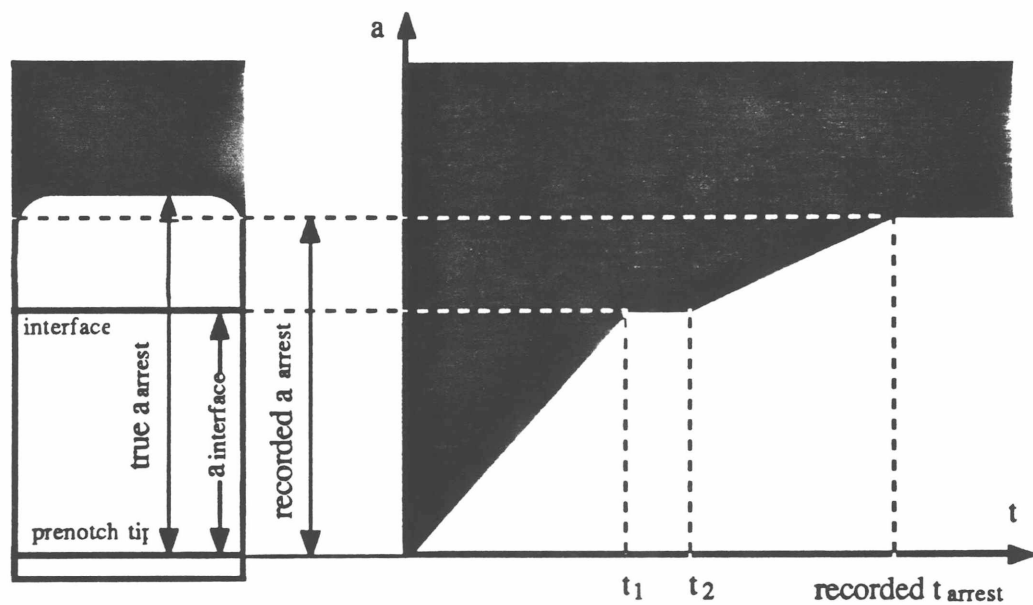


Figure 3.5.2.10: The crack front shape and the streak photography

d) In both the PMMA and the PC, uniform fracture surfaces were associated with constant speed RCPs (figure 3.5.2.11). When fracture surfaces exhibited non-uniform appearance (ripple marks, off-plane cracks etc.), streak mode records were largely obscured because the flash light was obstructed while passing through the crack gap of the fracturing specimen. Therefore, only tests with uniform fracture surfaces will be analysed (see chapters 5 and 6).

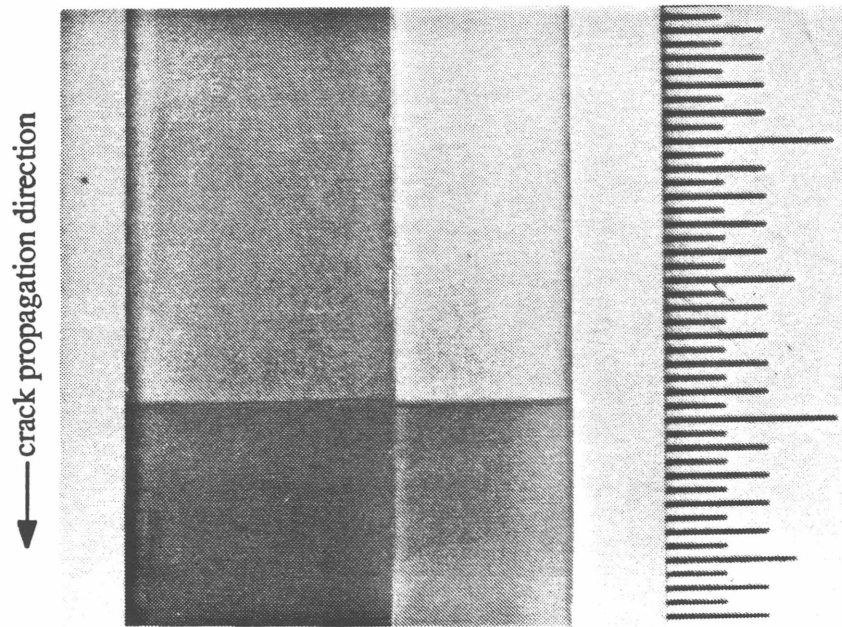


Figure 3.5.2.11: Fracture surfaces of completely broken samples (metric ruler)

- e) The minimum speed at which the crack propagated through the PMMA/PC interface was 280 m/s, while the maximum speed at which it arrested at the interface was 425 m/s. This is not in agreement with results published by Rolland[83, 85] (section 2.4.2). Testing the similar materials, he reported a minimum critical speed of 320 m/s above which the crack always penetrated into the PC part of the specimen, and below which it always arrested at the interface.
- f) Details of the crack length at initiation were not recorded and the initiation time will be extrapolated (section 5.3.2).
- g) The bluntness of the prenotch and the size of the specimen showed no clear effect on the crack behaviour in the PMMA.

h) Generally, a specimen with longer prenotch relative to the specimen width, failed at a lower breaking load. Figure 3.5.2.12 presents the load-displacement curves of two samples with the same dimensions and different lengths of prenotch.

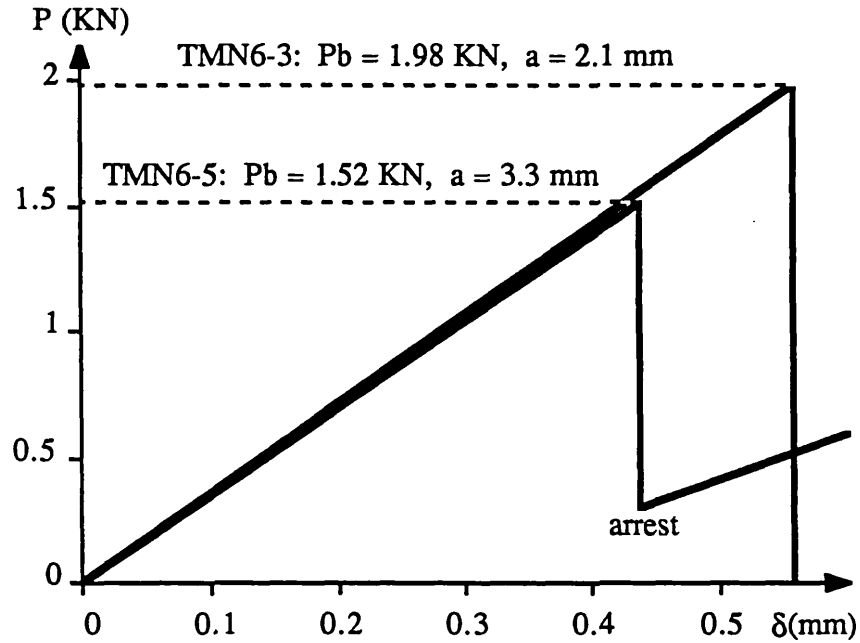


Figure 3.5.2.12: Typical load-displacement curves of DS with different initial notches

Specimens TMN6-3 and TMN6-5 (table 3.5.2.1) were provided with 2.1 mm and 3.3 mm long prenotches, respectively. During the testing, specimen TMN6-3 fractured completely, while crack arrest occurred in sample TMN6-5. The main reason for the crack arrest was the lower load at which the specimen with the deeper precrack failed. Although the crack reached the PMMA/PC interface, it did not have enough available elastic energy to be driven forward.

CHAPTER 4

EXPERIMENTS ON PMMA, PC AND PVC

4.1 Introduction

The previous chapter described the experimental study on the PMMA/PC two-layer system. In addition to this study, RCP experiments on both the PMMA and PC are performed. While RCP in the PMMA is studied by testing a SENT sample (section 4.2.1), a glass-duplex SENT sample is employed to generate a fast crack in the PC (section 4.2.2). Also, a limited number of experiments is done on the PVC by using both duplex PMMA/PVC and glass-PVC SENT specimens (sections 4.2.2-3).

The fast photography technique was employed for the recording of RCP in the duplex PMMA/PC SENT geometry (section 3.4). Unfortunately, this technique produced the appropriate RCP records for only 60% of all tested samples. Therefore, in the rest of this work, the electric grid line technique is used instead. After presenting the details of the crack speed measurements (section 4.3), the main results are summarized in section 4.4.

4.2 SENT specimen details

4.2.1 PMMA SENT sample

The SENT configuration is used throughout the experimental study of the PMMA (figure 4.2.1.1). The samples were cut from 13 mm thick PMMA sheets. The specimens were prenotched by machining an initial notch (a_0), followed by sliding of a new razor blade to produce a sharp notch (a_1). The total length of the prenotch ($a_0 + a_1$) was 20% of the specimen's width D (figure 4.2.1.1). Pushing the razor blade could lead to a blunt notch (Genussov[89]), and additionally, the depth of the resulting precrack might not be uniform (section 3.4). PMMA SENT samples with various width to height ratios (D/H) were examined: widths - $D = 10, 30, 60$ mm, were combined with three different heights - $H = 17, 40, 80$ mm. The reason for testing specimens of various dimensions was to investigate the influence of the specimen size on RCP properties of the tested material (chapter 5).

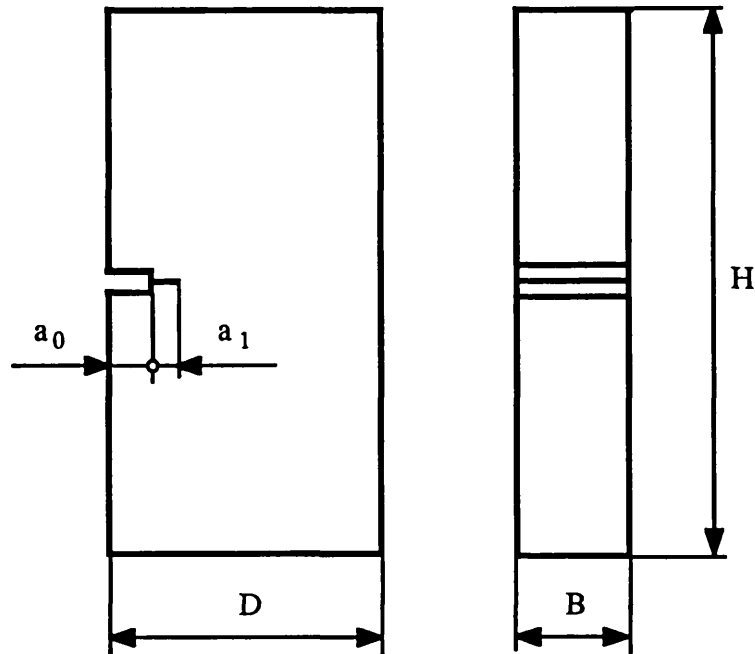


Figure 4.2.1.1: SENT specimen configuration

4.2.2 Glass-duplex SENT sample

The problems involved in monitoring and analysing impact tests, frequently used to introduce RCP in tough polymers, are related to gross oscillations of the sample, its slippage and friction on the supports, bouncing of the striker, contact stiffnesses, imprecise determination of boundary conditions of loading and deflection, etc. Genussov[89] overcame these difficulties by developing a glass-duplex SENT specimen (sections 2.2.2, 2.4.2). The main characteristic of this loading technique is that a sharp crack was initiated in the slowly preloaded sample that possessed enough stored strain energy to drive the fast crack forward. Using this technique, not only that RCP was successfully generated in tough polymers, but the modelling of RCP was considerably simplified and improved compared to modelling of RCP introduced by impact (Genussov [89]). The glass-duplex specimen consists of a glass slide bonded over the crack mouth of a sidegrooved SENT sample (figure 4.2.2.1). The sample is loaded slowly in tension until the glass breaks. Fracture of the slide is expected to occur at the maximum stressed cross section, which is right opposite the precrack. However, Genussov[89] claimed that when an unpolished glass slide was used, the large flaws introduced by the sawing process led to the slide breaking at arbitrary locations or prematurely at low loads. Thus, he suggested the use of a polished glass slide because it improved loading capacity and made the slide break at the desired location. Side-grooves were machined in order to obtain a straight crack path and to reduce free surface effects on the moving crack behaviour (Genussov [89]). While the sample is loaded slowly in tension, the glass-slide reduces the opening of the crack mouth because it carries most of the applied load (glass is much stiffer than the tested material). Consequently the crack tip was prevented from blunting. When the glass suddenly breaks, the sudden 'release' of the crack mouth generates shear waves. Those travel towards the still sharp crack tip, loading it at a high rate. In this way, it is possible for a brittle crack to initiate in a tough material (Genussov [89]). Since the glass-slide breaks at a high applied load,

the amount of energy stored in the sample should be sufficient to propagate a fast crack.

In the present work, the glass-duplex configuration was used for the analysis of RCP in PC and PVC materials (figure 4.2.2.1).

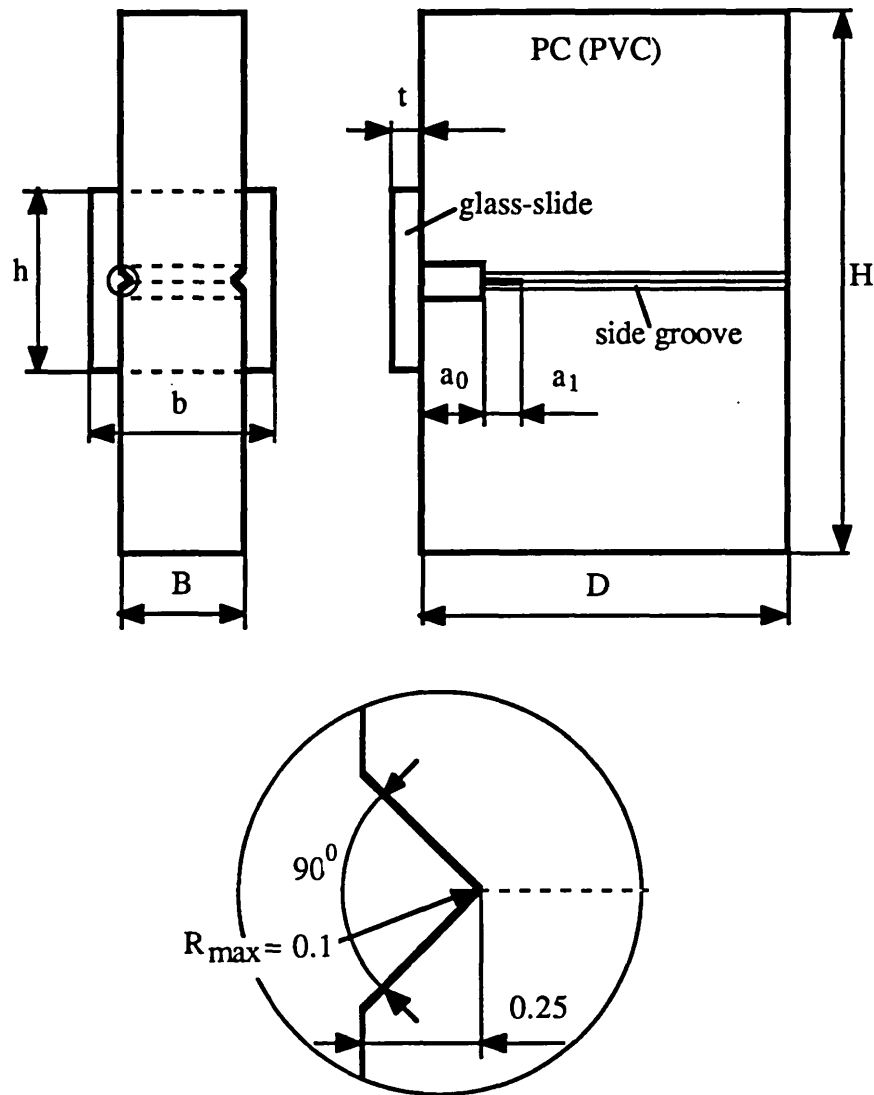


Figure 4.2.2.1: Glass-duplex sample configuration, dimensions in (mm)

The specimens were cut from 9.4 mm thick PC and 11.3 mm thick PVC sheets. Different width to height ratios (D (10, 30, 50 mm) / H (19, 30-40, 60-80 mm)) were employed for both glass-PC and glass-PVC SENT samples (figure 4.2.2.2, section 4.2.1). The specimens were

prenotched in two steps: milling of an initial notch (a_0), and then sliding of a fresh razor blade for a sharp notch (a_1). The total length of the prenotch ($a_0 + a_1$) was around $0.2D$. To avoid crack deviation from the axis of symmetry and encourage propagation of a straight crack front, a sharp, 0.25 mm deep grooves were machined on both sides of each sample (figure 4.2.2.1). In practice, difficulties were experienced in aligning the two side-grooves and the prenotch.

Glass-slides were sawn from a 3 mm thick sheet of an ordinary window soda lime glass (as recommended by Genussov[89]). The width b of the glass-slides was kept constant at 14 mm for all specimens (figure 4.2.2.1). The height h was varied between 10 and 20 mm according to the size of the sample (figure 4.2.2.2). The surface of the polished glass-slide was cleaned by acetone, while the corresponding surface of the polymer (either PC or PVC) was treated first by the sand-blasting technique and then with ethyl alcohol (section 3.2). An excellent adherence between the glass-slide and the PC (PVC) was achieved with Super Glue 3 (product of Loctite, UK).

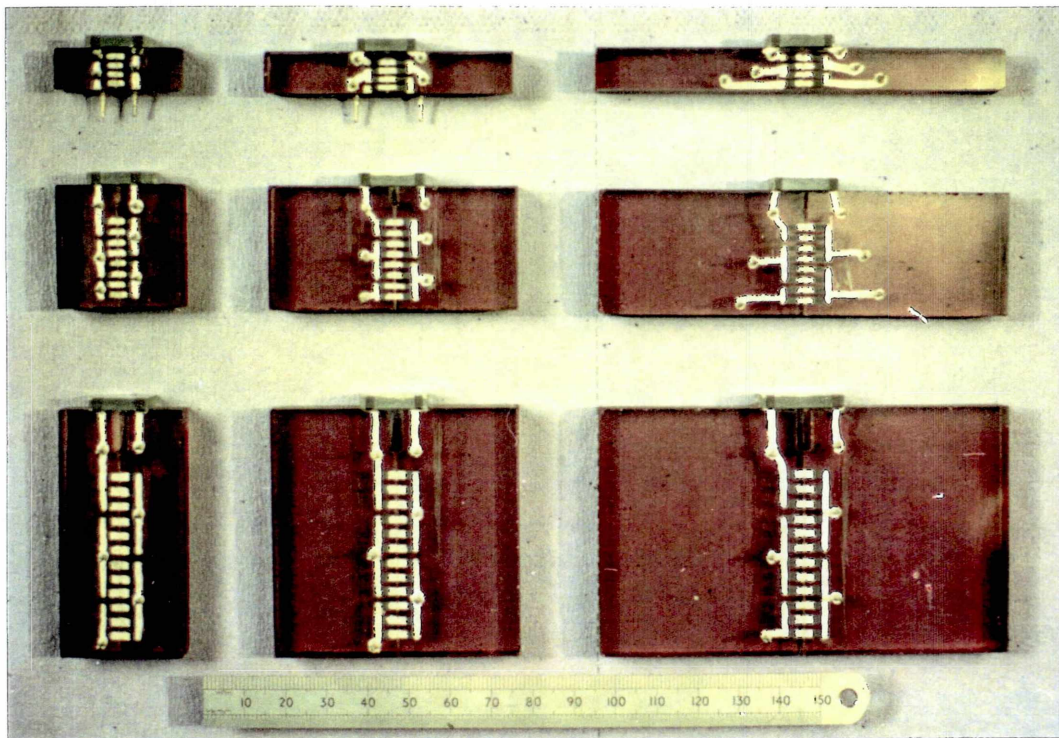


Figure 4.2.2.2: Glass-PVC SENT samples with various D/H

4.2.3 Duplex PMMA/PVC SENT sample

Due to project constraints, the tests were conducted on only ten duplex PMMA/PVC SENT specimens. The method of preparing the specimens was identical to that of the duplex PMMA/PC SENT samples, and is described in section 3.2. All samples had approximately the same dimensions: $D_1 * D_2 * B * H = 13. * 11.3 * 10 * 100$ mm (figure 3.2.1). The total length $a_0 + a_1$ of the initial prenotch was around 1 mm.

4.3 Setup and crack speed measurement

The testing was conducted at room temperature using a conventional Instron machine. The specimens were clamped between two heavy grips, and loaded slowly in tension at a loading rate of 10 mm/min. The load and the cross head displacement were recorded by the internal Instron chart recorder (section 3.4).

A resistance timing-line method (section 2.3.2) was employed to monitor the crack speed in the samples described in the previous section. A graphite paint (Electrodag 6012, Anti-static coating, product of Acheson Colloids Company, UK) was sprayed on the samples through plastic stencils. In this way, a series of parallel conducting lines was produced transverse to the prospective crack path. Depending on the width D of the samples (figures 3.2.1, 4.2.1.1, 4.2.2.1), a number of conducting lines was: 4 for $D = 10$ mm, 8 for $D = 30$ mm and duplex PMMA/PVC samples, and 12 for $D = 50$ -60 mm (figures 4.2.2.2, 4.3.1-3). The thickness of the lines was 1 mm for 4 and 8 line patterns, and 1.5 mm for 12 line pattern. The distance between the lines was 1, 1.5 and 2 mm for 4, 8 and 12 conducting line patterns respectively.

Before spraying the graphite paint, an adhesive tape approximately 3 mm wide was adhered to the sample. Thus, a thin strip of the specimen surface around the crack path was not sprayed. Short silver paint lines (section 3.4) were then painted on the unsprayed surface, connecting the corresponding parts of graphite lines (figures 4.3.1-3). The reason for painting the silver lines across the crack path is that continuous graphite lines would result in a 'strain-gauging' effect (Genussov[89]). Because of large strains near the crack tip, the resistivity of the continuous graphite line changes gradually before the line is broken. Consequently, a sharp step in voltage reading cannot be obtained. The timing-line technique is an incremental crack monitoring technique. An average crack speed between the timing lines can be obtained from the distance between adjacent lines and the time interval between corresponding voltage steps. Thus, a sharper change in the voltage reading results in a more accurate evaluation of the crack speed.

For improved resolution of the resistance step reading, the timing lines were connected in four groups using silver paint (Genussov[89], section 2.3.2, figures 4.2.2.2, 4.3.1-3). In the case of the PMMA and duplex PMMA/PVC SENT samples, an additional silver paint line was painted in front of the tip of the precrack for the triggering purpose (figures 4.3.1-2). A triggering pattern, consisting of two silver lines painted on the glass-slide and connected in parallel, was employed for the glass-PC and glass-PVC SENT specimens (section 3.4, figure 4.3.3). Six 'eyelet-type' metal terminals (connection pins) were placed at the end of the conducting lines (figure 4.2.2.2). A Nicolet transient recorder (model 4096) was used to monitor the step changes in the voltage as the timing lines were cut by the propagating crack. The oscilloscope was set on its fastest sampling rate (resolution of 0.5 μ s per point). It was set to store 8 ms of data, starting 2 ms before triggering. Both the triggering and timing lines signals were displayed on the screen and stored on a floppy disc. The accuracy of the speed measurement can also be increased by increasing the number of the timing lines (section 2.3.2). This was not necessary in the present work, because rather constant crack speeds were observed.

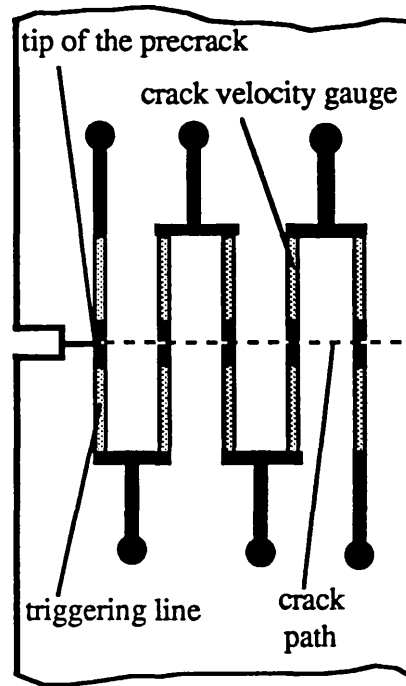


Figure 4.3.1: Timing line configuration, 4 line pattern on PMMA SENT sample, (solid - silver paint, dotted - graphite)

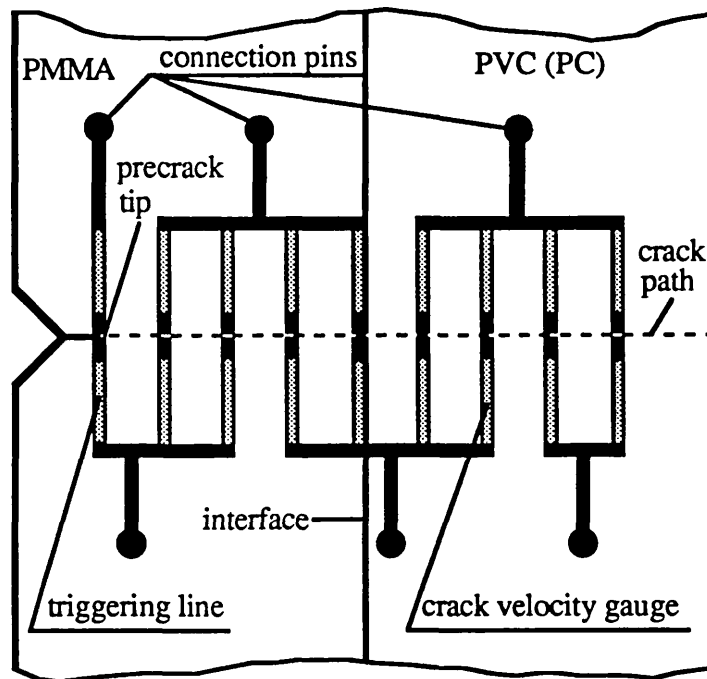


Figure 4.3.2: Timing line configuration, 8 line pattern on PMMA/PVC SENT sample, (solid - silver paint, dotted - graphite)

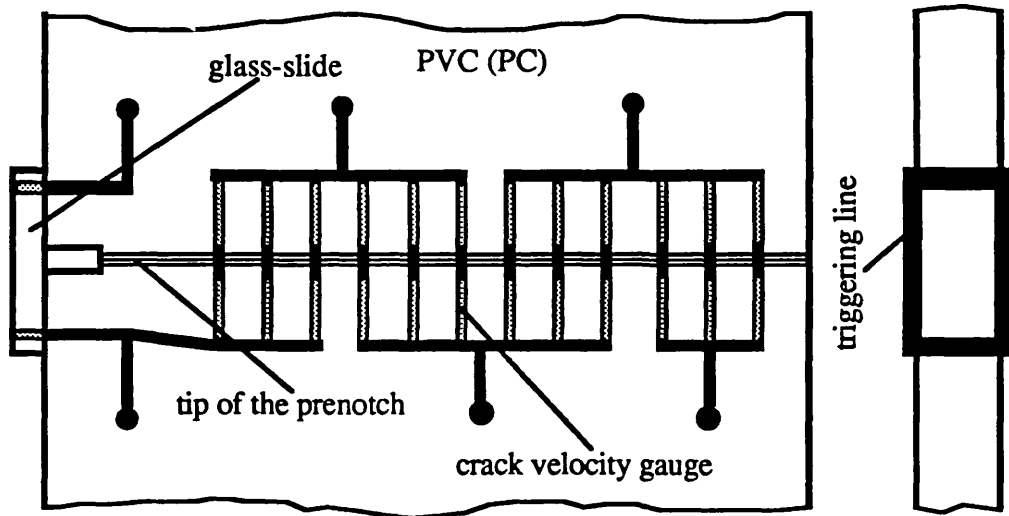


Figure 4.3.3: Timing line configuration, 12 line pattern on glass-duplex SENT sample, (solid - silver paint, dotted - graphite)

4.4 Results

4.4.1 PMMA SENT results

About 90% of all tested PMMA SENT samples generated satisfactory results - planar uniform fracture surface and 'clear' record of the crack history data. This is a much higher percentage of successful tests than achieved using the high speed photography for recording of RCP in the duplex PMMA/PC samples (section 3.5.2). The timing-lines technique involves a longer preparation of the sample than the high speed photography method, but the former technique showed to be more reliable.

Typical records of the step changes in voltage during RCP in PMMA SENT samples are presented in figures 4.4.1.1-2.

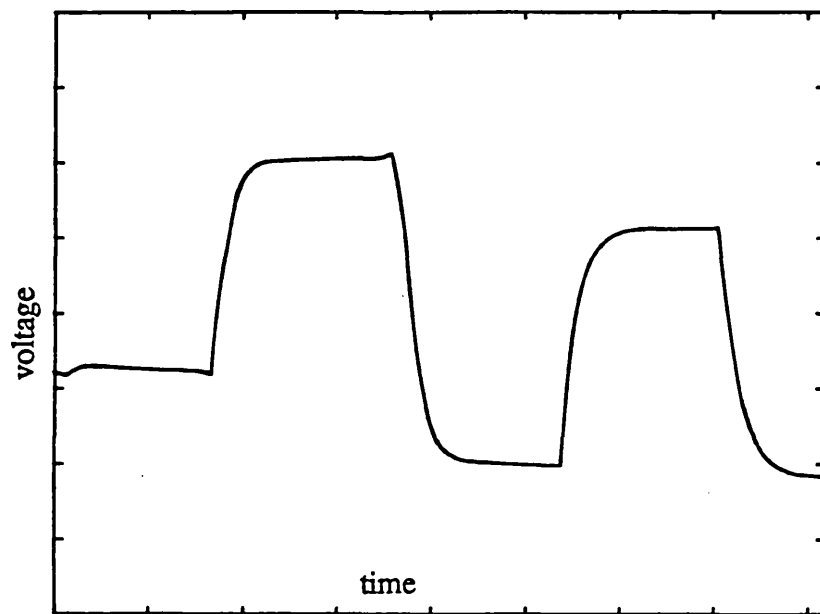


Figure 4.4.1.1: Voltage vs time record from 4 line pattern, test PM4-4

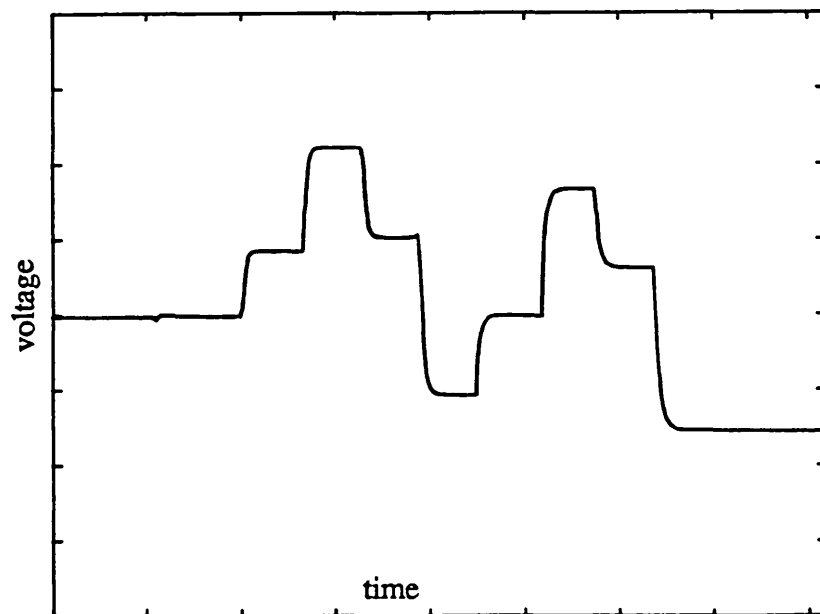


Figure 4.4.1.2: Voltage vs time record from 8 line pattern, test PM5-1

The most important observations concerning RCP experimental results of the PMMA SENT samples are summarized as follows:

- a) All 'clear' records of the crack history data showed a remarkably constant speed crack propagation (figures 4.4.1.1-2). Recorded crack speeds were in the range of around 200 to 420 m/s.
- b) Uniform fracture surfaces were associated with these constant speed RCPs (figure 4.4.1.3).

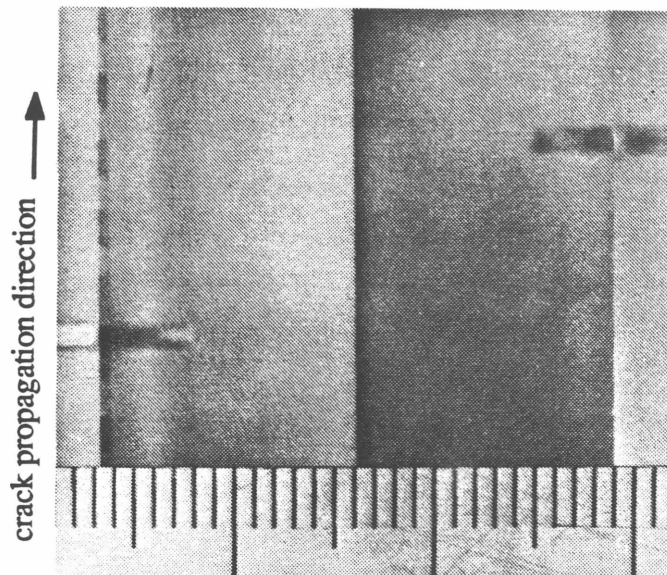


Figure 4.4.1.3: PMMA fracture surface, $\dot{a} \approx 230$ m/s, (metric ruler)

It should be mentioned that surface roughness rises with crack speed (figures 4.4.1.4-5). This has been observed by a number of authors. Cotterell[65] reported a smooth glassy surface at moderate crack speeds (less than 200 m/s). As speed increased, parabolic marks of increasing density appeared. He speculated that those marks (see figure 4.4.1.5) were due to micro-fracture from flaws ahead of the main crack, followed by a rough region (rippled surface) with small off-plane cracks. This was claimed to be due to microcracks suddenly coalescing,

causing a crack jump. Rosenfield and Kanninen [73] speculated that the parabolic marks might be caused by local heating, from the energy of fracture causing polymer decomposition and gas release forcing the crack apart. They reported that the rippled surface occurred at around 400 m/s with branching at 700 m/s. Crouch[86] observed mirror smooth surface of PMMA at speeds below 200 m/s. Parabolic marks of increasing density were reported to appear as speed increases, giving a matt uniform surface appearance. Crouch[86] also observed rippled marks across the specimen thickness. He obtained a maximum speed of 700 m/s from tensile tests, and reported that branching of the crack into multiple fracture occurred after gross surface roughening caused by failed off-plane cracks. Green[71], and Green and Pratt [74] observed the parabolic marks on the PMMA fracture surfaces at speeds between 200 to 400 m/s. They described those marks as discontinuous crazes growing from flaws ahead of the main crack along the fracture plane. The rippled surface, observed at high speed (around 500 m/s) and applied loads, was explained as similar discontinuous crazing which began to occur off the main fracture plane and coalesced together by a shear process. Recently, Shioya and Ishida [91] speculated that the parabolic marks are the intersections of the main crack front and the microcracks which nucleate, grow and coalesce into the main crack. At higher crack speeds, the fracture surface was reported to turn to a rough crazing surface with bifurcation attempts.

c) Higher crack speeds were found to correspond to higher breaking loads - P_b (see sections 3.5.2, 5.3.2).

d) Since the timing line technique is an incremental technique of monitoring the crack history data, the details of the crack behaviour at initiation were not recorded and the initiation time will be extrapolated (see sections 5.3.2-3). However, the crack propagation speed between the triggering line and the nearest timing line was remarkably close to the speeds recorded along the rest of the crack path. This suggests constant speed RCP throughout.

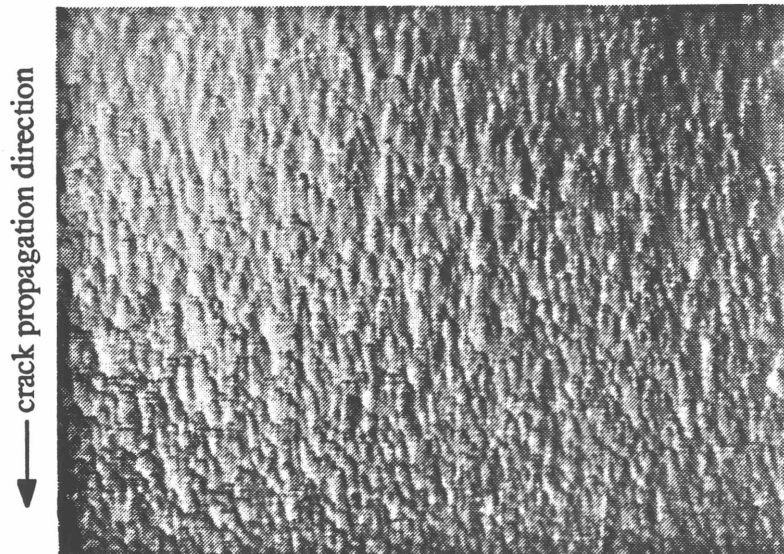


Figure 4.4.1.4: PMMA fracture surface, $\dot{a} \approx 200$ m/s (magnification x17.2)

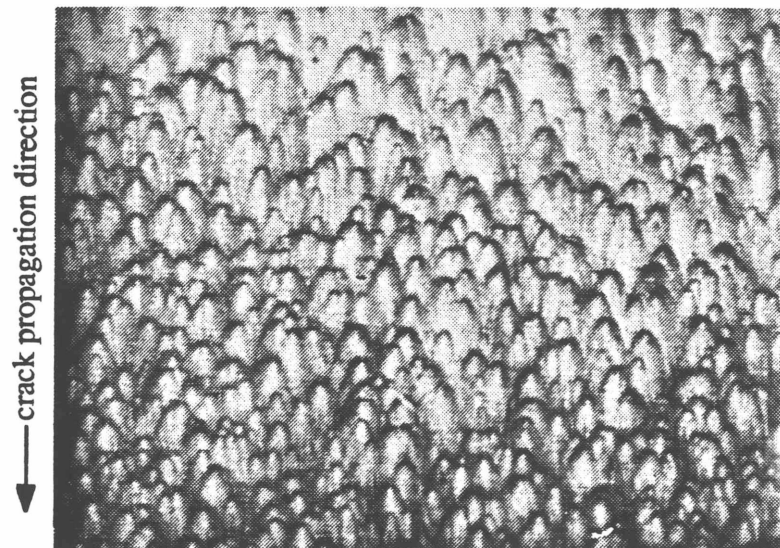


Figure 4.4.1.5: PMMA fracture surface, $\dot{a} \approx 380$ m/s (magnification x17.2)

e) One set of the PMMA SENT partially gripped samples was tested as will be explained in detail in section 6.2.1. Typical voltage vs time records obtained from those samples are presented in figures 4.4.1.6-7. Both figures show that the crack initially propagated at a constant speed over the first six intervals of the resistance grid. Beyond these six intervals, a reduction in crack speed was observed. In some cases, even crack arrests and reinitiations were observed, e.g. the crack arrested between the tenth and the eleventh timing line of the sample whose voltage vs time record is presented in figure 4.4.1.7. The changes in fracture surface appearances along the crack path corresponded to the variations of the crack speed.

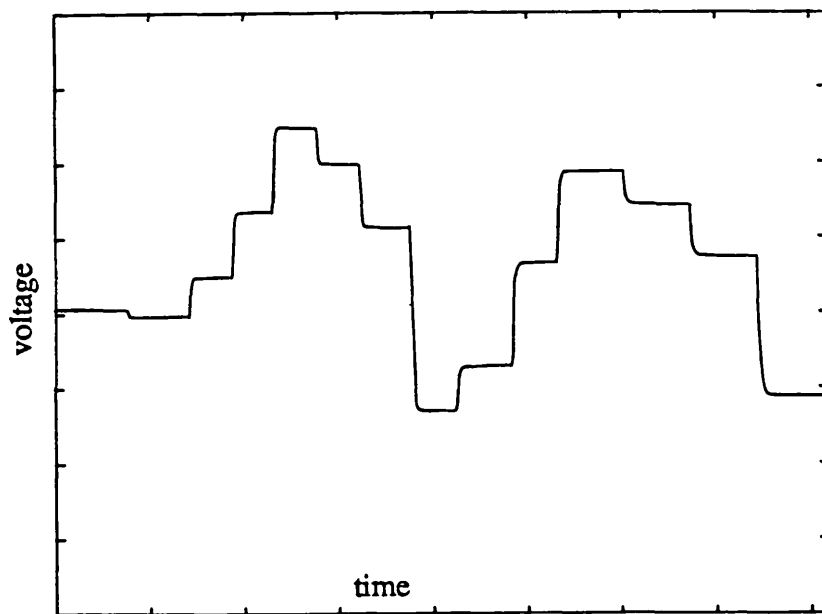


Figure 4.4.1.6: Voltage vs time record of non-uniform speed RCP (test PM9-4, section 6.2.1)

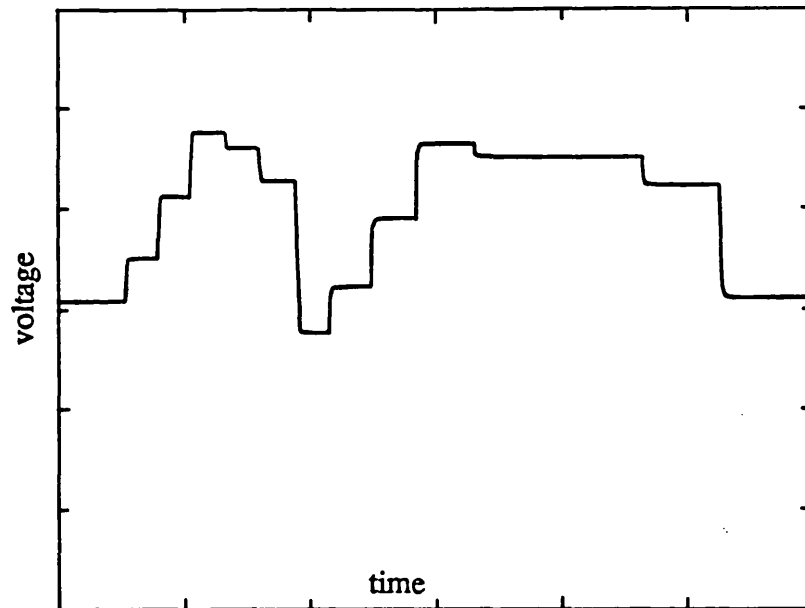


Figure 4.4.1.7: Voltage vs time record of non-uniform speed RCP (test PM6-4, section 6.2.1)

4.4.2 Glass-PC SENT results

It was mentioned in section 3.2 that the accuracy of G_{ID} results generated by the indirect method depends on the accuracy of the monitored crack tip position during RCP. In order to monitor the exact crack tip location by the surface technique, the crack should be encouraged to propagate with a straight front. Furthermore, the dynamic two-dimensional (2-D) FE code, used in this work, can only model the crack which advances with a straight front along the planar path. Thus, only RCPs associated with uniform planar fracture surfaces will be analysed numerically (chapter 5).

A total of about 50 glass-PC SENT samples were tested. Unfortunately, only about 40% of the tested samples generated the required mode of failure (uniform planar fracture surface) and the appropriate record of the crack history data. The main problem involved in using the glass-duplex configuration for testing of RCP in the PC was the unpredictably high applied load at which the glass broke. When this load was excessively high, the crack initiated not only from the tip of the prenotch but from the side-grooves along the reduced net section of the sample as well. Furthermore, even the slightest error in aligning the two side-grooves and the prenotch led to crack deviation from the planar surface and gross roughening of the fracture surface (figure 4.4.2.1). These anomalies in the crack behaviour resulted in irregular voltage vs time records (figures 4.4.2.2-3), and were especially pronounced in wider specimens. Sometimes, gross yielding occurred followed by fracture at the loaded boundary before the glass-slide broke.

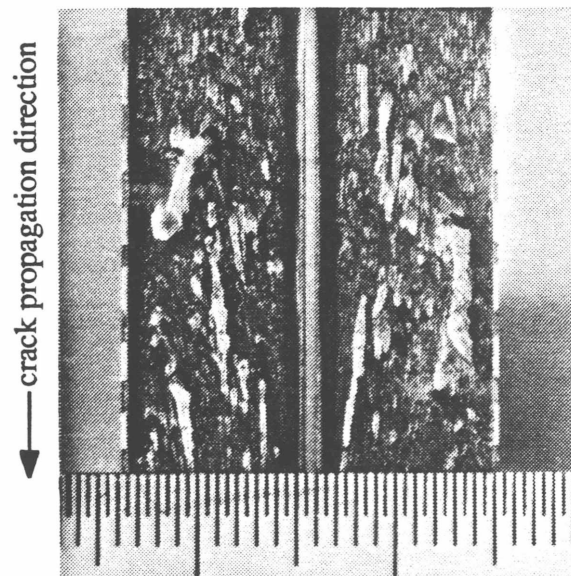


Figure 4.4.2.1: Gross roughening of PC fracture surface, $\dot{a} \approx 450$ m/s
(metric ruler)

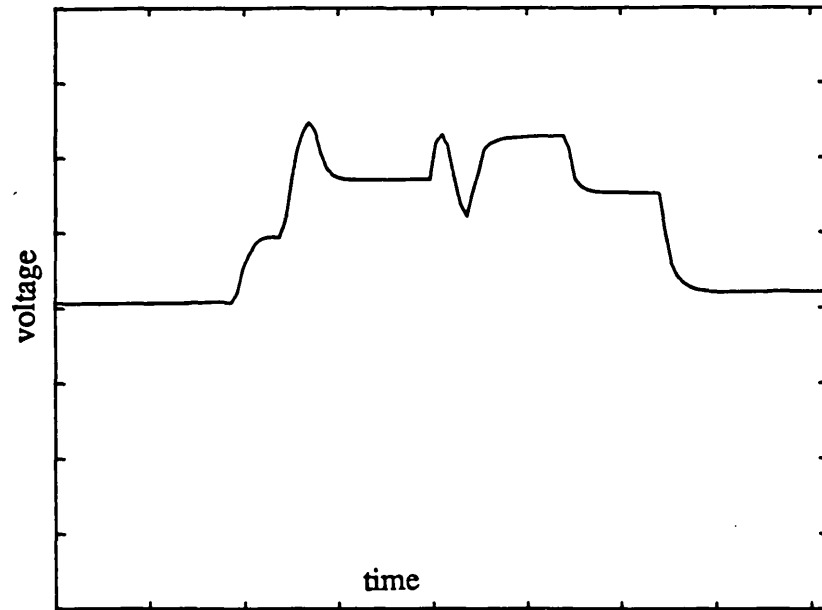


Figure 4.4.2.2: Irregular voltage vs time record from 8 line pattern

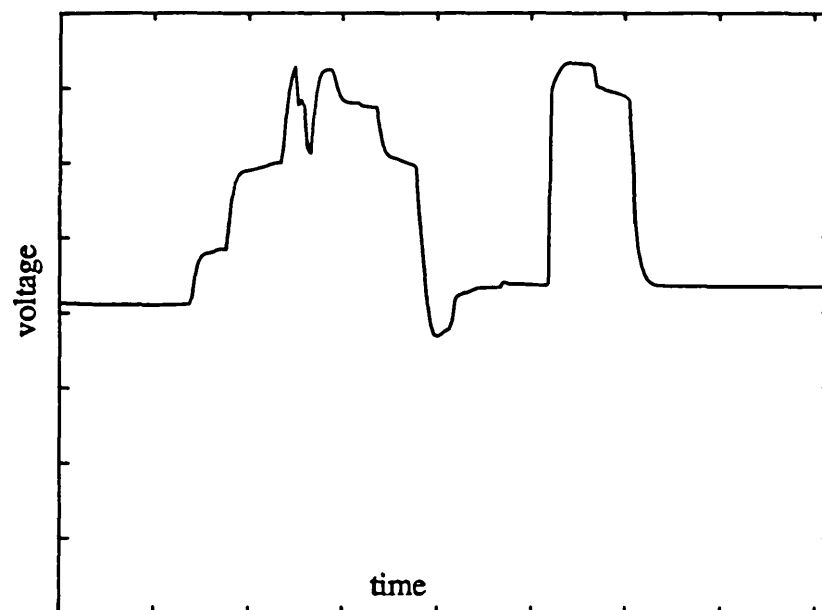


Figure 4.4.2.3: Irregular voltage vs time record from 12 line pattern

Most of the tests on glass-PC samples, which generated planar fracture surfaces and 'clear' voltage vs time records, resulted in a crack speed of about 500 m/s. Only a few crack speed results were around 400 m/s. As in the PMMA SENT samples, the crack propagated at remarkably constant speed leaving a uniform fracture surface behind (figure 4.4.2.4). Also, higher crack speeds were associated with higher loads at which the glass-slides broke. In a number of the samples crack branching attempts as well as successful crack branching into a multiple fracture were observed (figure 4.4.2.5). Failed branch attempts were not necessarily fully through-thickness. The lowest speed at which crack branching attempts were observed was 475 m/s, while the highest speed at which the crack propagated without bifurcation was 648 m/s. Crack branching was always associated with high breaking load and high crack speed. This observation is in accordance with the results of Takahashi and Arakawa [90]. They claimed that the product of the crack speed and crack driving force must exceed a certain unique value in order for the crack to bifurcate. A thorough review of crack branching theories is given by Crouch[86], Ramulu and Kobayashi [85], Ravi-Chandar and Knauss [84-b] etc.

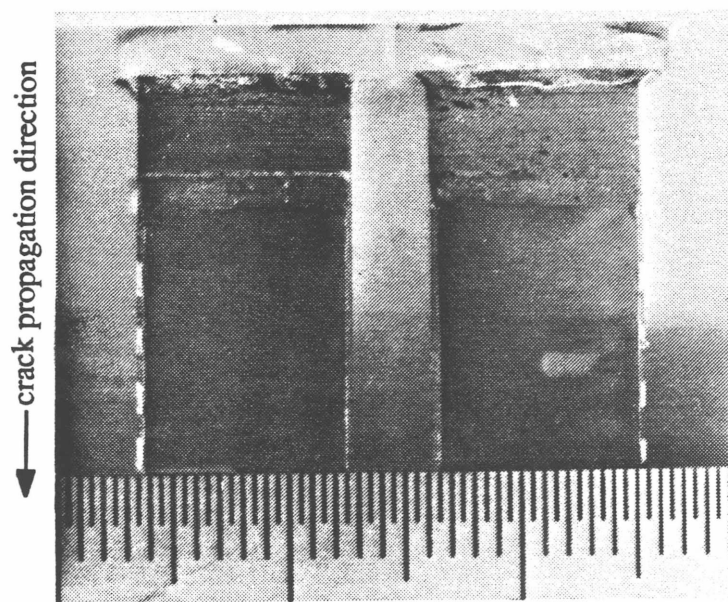


Figure 4.4.2.4: Uniform fracture surface of PC, $\dot{a} = 437.5$ m/s
(metric ruler)

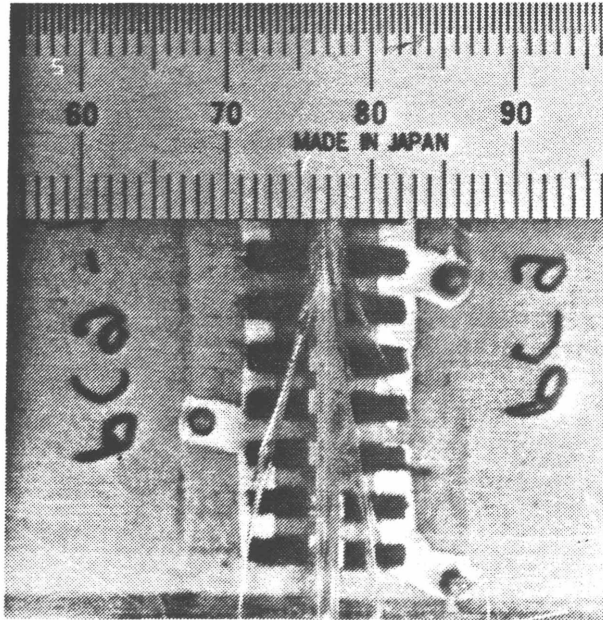


Figure 4.4.2.5: Multiple crack branching in PC, $\dot{a} \approx 550$ m/s

The details concerning the crack behaviour at initiation were not recorded (sections 2.3.2, 4.4.1), and the initiation time will be extrapolated according to the subsequent crack speed in PC and the time difference T_{AC} between the triggering instant and the first step change in the voltage vs time record (figure 4.4.2.6, see section 5.3.4). The value of T_{AC} varied between 14 to 30 μ s. Larger T_{AC} corresponded to longer distances between the triggering line and the first timing line - 6.5 to 20.5 mm. No clear dependence of T_{AC} on the crack speed was observed. The accuracy of the measured T_{AC} was solely determined by the quality of the voltage vs time record, i.e. by the sharpness of the voltage changes. Since the oscilloscope was set on its fastest rate (the resolution of 0.5 μ s per point was used) and these changes were usually very sudden, the error in monitoring of T_{AC} should not be higher than 2 μ s.

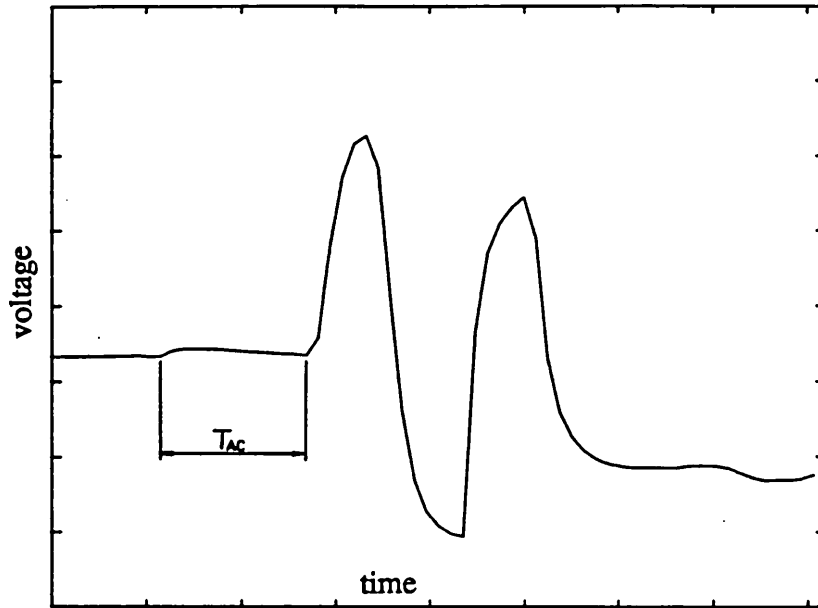


Figure 4.4.2.6: Voltage vs time record from 4 line pattern

Crouch[86] employed the SENT geometry for testing of RCP in PC and PVC. When the samples were loaded slowly at RT, nominally sharp notches blunted in both PC and PVC samples. This often led to branching of the crack, and only a limited range of speed data were obtained (about 450 m/s in PC). Therefore, he tested precooled SENT samples. However, obtained crack speed results showed that the crack propagated at constant speed. In PC, Crouch[86] reported a limited number of smooth fracture surfaces followed by uniformly matt features. Visible rippled lines similar to those of PMMA were observed on PC fracture surfaces (Crouch[86], Glover et al [73]). Also, some river lines and longitudinal marks were reported. Doyle[83] noted smooth/rough lines (rippled marks) on the surface of polycarbonate. He attributed these to the sudden fracture of a growing craze and subsequent repetition of the craze growth.

4.4.3 Glass-PVC SENT results

The glass-duplex configuration was used to generate RCP in PVC. Unfortunately, all crack behaviour anomalies observed in the PC specimens (see previous section), were observed in the PVC specimens as well. Although a number of tested samples resulted in reasonably 'clear' records of the crack history data, none of them generated the required mode of failure (uniform planar fracture surface). Consequently, numerical analysis of the experimental RCP results of PVC was not performed (sections 3.2, 4.4.2). No RCP at a crack speed below 500 m/s was obtained. About 90% of the specimens fractured at a crack speed greater than 550 m/s. Below this speed, extremely rough fracture surfaces were observed. At a speed of 550 m/s and above, RCP in glass-PVC samples was always associated with crack branching (figure 4.4.3.1). Obviously, the glass-duplex loading technique cannot be successfully employed for testing of RCP in PVC, because of excessively high loads required for the glass-slide to break (section 4.4.2). Crouch[86] reported that slowly loaded PVC SENT samples tended to yield and that brittle fractures were initiated at -20° C or below (see previous section). He observed uniformly matt fracture surfaces at low crack speeds, while rougher surfaces, consisting of random fluctuations of increasing severity, were associated with higher speeds. Crack branching was claimed to be unpredictable with no apparent definite angle for branching.

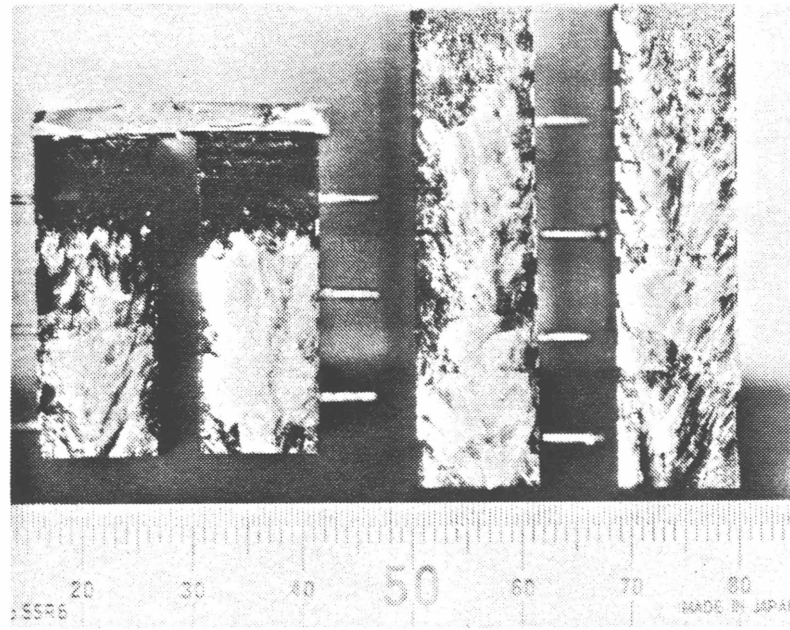


Figure 4.4.3.1: Crack branching in PVC

4.4.4 Duplex PMMA/PVC SENT results

The fast photography method was employed to obtain a continuous record of the discontinuous crack behaviour in the duplex PMMA/PC SENT samples (sections 3.4, 3.5.2). However, due to its simplicity, the resistance timing-line technique was used for recording of RCP in the duplex PMMA/PVC samples (section 4.3, figure 4.3.2). This technique was first applied to the duplex PMMA/PC samples, and obtained incremental RCP results were compared against the continuous results of section 3.5.2. The timing-line technique successfully detected most of the features of RCP in the PMMA/PC geometry, previously observed by the fast photography method (section 3.5.2). Thus, the applicability of the method seems to be validated even for the geometries where discontinuous crack behaviour might be expected.

The main observations regarding the results obtained from the duplex PMMA/PVC SENT samples are summarized below:

- a) The crack propagated at a constant speed in the PMMA part of the sample, leaving a uniform fracture surface behind (figure 4.4.4.1).
- b) The crack crossed the interface and entered the PVC part of the PMMA/PVC specimen with no apparent crack 'hesitation' at the interface (unlike in the PMMA/PC samples, section 3.5.2).
- c) RCP was characterized by a higher crack speed in the PVC than in the PMMA part. The recorded crack speed was in the range of 330 to 450 m/s in the PMMA and 570 to 630 m/s in the PVC part of the duplex specimen.
- d) The crack branching always occurred in the PVC part of the sample (figure 4.4.4.1).

The results of the tests on both the glass-PVC and the PMMA/PVC configurations, showed that even under slow loading conditions at RT, PVC may become very susceptible to RCP.

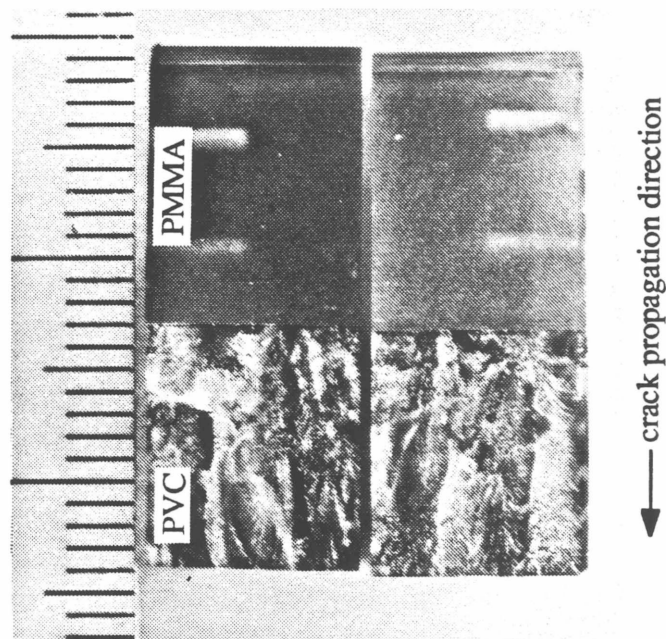


Figure 4.4.4.1: PMMA/PVC sample, typical fracture surface appearances (metric ruler)

CHAPTER 5

NUMERICAL ANALYSIS OF EXPERIMENTAL RESULTS

5.1 Introduction

The basic problem in the analysis of a body containing a propagating crack is to solve the equation of motion:

$$\text{div}[\sigma] + \rho\{b\} - \rho \, d\{v\}/dt = 0 \quad 5.1.1$$

$[\sigma]$ - stress tensor, ρ - mass per unit volume, $\{b\}$ - body forces per unit mass,
 $\{v\}$ - velocity field vector.

As the constitutive equations of linear elastic material behaviour are introduced into the equation of motion (equation 5.1.1), all stress components are expressed in terms of strain components, and hence in terms of displacement derivatives. Having obtained the displacements and velocities throughout the sample at any instant of time, the global energy balance equation (equation 1.2.2.2) together with a crack extension criterion (most often in the form of equation 1.2.2.6), can be employed to determine either the crack speed, if RCP properties (usually K_{II} or G_{II}) are known, or RCP properties if the crack velocity is known.

Given the analytical difficulties described in section 1.2.2, some form of numerical solution is necessary. This chapter briefly discusses general basis of dynamic Finite Element (FE) method of modelling RCP, describes the FE program used in the analysis of experimental results from chapters 3 and 4, and presents the results of the elastic generation-mode simulations.

5.2 FE modelling of RCP

5.2.1 General basis

A number of numerical schemes have been developed in order to solve RCP problems. These schemes utilize conventional numerical tools such as finite difference, finite element and boundary element. In this work, the FE technique is used to model RCP and is briefly described below.

FE representation of a continuous media consists of a mesh of a finite number of elements. Using the principle of virtual work, the well known FE matrix equation of motion is derived as (Zienkiewicz[71]):

$$[K] \{\delta\} + [M] \{\ddot{\delta}\} = \{F\} \quad 5.2.1.1$$

$[K]$ - the stiffness matrix of the whole structure,

$[M]$ - the mass matrix of the whole structure,

$\{F\}$ - external force vector,

$\{\delta\}$ - the displacement vector whose elements represent the displacements of all nodal points of the structure,

$\{\ddot{\delta}\}$ - the acceleration vector whose elements represent the accelerations of all nodal points of the structure.

There are two main procedures for RCP modelling:

- a) The fixed mesh - 'node release' procedure, and
- b) The moving mesh - 'moving singular element' procedure.

Because of its simplicity (Crouch[86], Keegstra[77], Genussov[89]), the fixed mesh procedure has been used in the present work, and will be discussed in the next section.

The dynamic FE simulations of RCP may be divided into two categories: generation and propagation modes. A generation mode is one where the crack length history is defined and the corresponding G^{dyn} (K^{dyn}) is generated. Thus, by assuming the equivalence of G^{dyn} and G^I during the crack propagation (equation 1.2.2.6), the dynamic crack resistance property of the material is obtained. The propagation mode predicts the propagating crack history. This scheme involves the specifications of an assumed G^I vs crack speed relationship, and a crack propagation criterion (equation 1.2.2.6). Both modes require the specification of the geometry, loading conditions and material mechanical properties. The propagation mode is more complex to program (Crouch[86]), and the form of G^I vs $\dot{a}$ curve is generally not known in advance.

5.2.2 Simulation of the propagating crack

When the fixed mesh technique is employed, the running crack is simulated by progressively changing boundary conditions of nodes along the crack path. At the moment the simulated crack tip reaches a node, the nodal boundary conditions change from constrained to free conditions. A step change of boundary conditions with instantaneous increase of the crack length by one nodal spacing tends to produce spurious oscillations in the solution, and there is no net energy loss from the system as the crack propagates. This problem can be overcome by applying a 'holding back force' to the crack tip node at the instant of release. Keegstra[77], for the propagating mode, and Yagawa et al [77], for the generation mode, suggested the concept of the 'holding back force' acting on the released node and decaying gradually from its maximum value equal to the nodal force prior to release. Thus, artificial oscillations were

removed and an energy sink near the moving crack tip was produced. Keegstra[77] showed that the crack tip nodal force at the instant of release was proportional to K^{dyn} , and so he used a critical tip nodal force dynamic propagation criterion. He decayed his holding back forces on a time basis until a critical crack opening displacement (COD) was reached. Consequently, more than one crack path node was held at any instant. Crouch[86] found difficult to justify the restraint of more than one node at any time given the coarse meshes used by Keegstra (the minimum nodal space was 2 mm).

Other fixed mesh FE studies have decayed holding back forces over one nodal spacing. Thus only one node was held at any instant. The holding back force, applied on the last released node, decayed gradually according to the exact position of the crack tip between the last node that was released and the fixed node ahead. A number of decay schemes have been proposed. These have the general form:

$$F_h = -F_o \left(1 - \frac{d}{\Delta}\right)^s \quad 5.2.2.1$$

F_h - holding back force, F_o - maximum force at the node before it was released, Δ - spacing between the released and fixed node, d - incremental distance the crack has travelled between the nodes.

The holding back force relaxed completely as the crack travelled the distance Δ , i.e. as the next node was released. The rate of the decay was governed by the parameter s . Rydholm et al [78] and Malluck and King[78] derived the two different values of s : 0.5 and 1.5 respectively. However, Malluck and King[80], Crouch[86] and Genussov[89] demonstrated insensitivity of the calculated energy release rate to the s parameter. Hodulak, Kobayashi and Emery[80] used Kalthoff's experimental K^{dyn} vs $\dot{a}$ data on DCB samples of Araldite as a bench mark, to examine a linearly decreasing holding back force ($s=1$). They predicted a crack history by using K^{dyn} as an input for the propagation mode, and calculated K_{ID} by employing experimental crack length history data as an input in the generation mode. A very good

agreement between experimental K^{dyn} vs $\dot{a}$ data and calculated values of both modes was reported. Genussov[89] employed a linearly decaying holding back force ($s=1$) in his analysis of glass-PE specimens (sections 2.2.2, 2.4.2). In the present work, the value of $s=1$ is used throughout.

5.2.3 Details of the FE program used in this study

The analysis of RCP in this work is based upon a dynamic FE code originally developed by Keegstra[77]. It is a fixed mesh - 'node release' technique of RCP modelling. Keegstra[77] compared his program in propagation mode to published experimental results on the DCB geometry (Hahn et al [75], Kobayashi et al [76], Kalthoff et al [77]), the CT geometry (Johnson et al [74], Fuller[77]) and the Charpy impact geometry (Glover et al [76]). Good agreement was reported. Keegstra's program was modified by Crouch[86] who implemented the conventional holding back force decay scheme (equation 5.2.2.1). Crouch[86] reported good agreement between his dynamic FE generation mode results and Kalthoff's [76, 83] experimental data on Araldite, as well as PMMA results published by Doll[76], Cotterell[65] and Rosenfield and Kanninen [73]. He also tested the model against Broberg's analytical solution (section 1.2.2) and showed excellent agreement. Recently, Genussov[89] modified Crouch's version in order to model viscoelastic (VE) response of polyethylene (PE) during RCP. He employed a VE model called 'Phases model' which is a modified version of the generalized Maxwell model. The model consists of several Maxwell elements in parallel, with non-linear dashpots. Each element is called a 'phase', and the sum of the responses of the different phases simulates the gross response of the polymer.

In this work, Crouch's version of Keegstra's program is used and is adapted for the generation mode analysis of RCP. The program was designed to analyse symmetrical bodies of isotropic elastic material. Due to symmetry, only half of the sample is simulated and the crack path becomes a boundary. Constant strain triangular elements are used in this 2-D code. Prior to the crack initiation, an initial static solution of the stress-strain distribution, corresponding to the initial boundary loads and displacements, is computed by using a conventional Gauss-Seidel iterative scheme. Having obtained the initial static solution, the crack is propagated by the sequential release of boundary nodes at times specified by the experimental data. During the crack propagation, the equation of motion (equation 5.2.1.1) must be repeatedly solved. The dynamic solution was computed by an iterative step by step implicit procedure (Zeinkiewicz[71]), as suitable for dynamic non-linear problems in which the matrices [M] and [K] are varying in time. The lumped mass matrix [M] is diagonal and is preferred for dynamic solutions since the equations of motion are uncoupled, simplifying the solution process.

A polynomial is assumed to relate the accelerations ($\{\ddot{\delta}\}$) at a series of sequential time values:

$$\begin{aligned} \{\ddot{\delta}(\tau)\} = & -\frac{1}{2}(1-\tau)\tau\{\ddot{\delta}\}_- + (1-\tau)(1+\tau)\{\ddot{\delta}\}_0 \\ & + \frac{1}{2}(1+\tau)\tau\{\ddot{\delta}\}_+ \end{aligned} \quad 5.2.3.1$$

where $\tau = (t - t_0)/\Delta t$, t is time and Δt is the magnitude of the time-step. Subscript '-' stands for the previous time-step, subscript '0' stands for the present time-step and subscript '+' stands for the future time-step. This may be differentiated twice at $t = t_0$ ($\tau=0$), to give expressions for $\{\ddot{\delta}\}_0$ and $\{\ddot{\delta}\}_0$. A Taylor series expansion is used to express the displacement and the velocity vectors at the future time step. This series expansion predicts:

$$\{\delta\}_+ = \{\delta\}_0 + \Delta t \{\dot{\delta}\}_0 + \frac{\Delta t^2}{2} \{\ddot{\delta}\}_0 + \frac{\Delta t^3}{6} \{\ddot{\delta}\}_0 + \frac{\Delta t^4}{24} \{\ddot{\delta}\}_0 + \dots \quad 5.2.3.2$$

and similarly for $\{\dot{\delta}\}_+$. Substituting for $\{\ddot{\delta}\}_0$ and $\{\ddot{\delta}\}_0$ gives:

$$\{\delta\}_+ = \{\delta\}_o + \Delta t \{\dot{\delta}\}_o - \frac{\Delta t^2}{24} \{\ddot{\delta}\}_- + \frac{5\Delta t^2}{12} \{\ddot{\delta}\}_o + \frac{\Delta t^2}{8} \{\ddot{\delta}\}_+$$

$$\{\dot{\delta}\}_+ = \{\dot{\delta}\}_o - \frac{\Delta t}{12} \{\ddot{\delta}\}_- + \frac{2\Delta t}{3} \{\ddot{\delta}\}_o + \frac{5\Delta t}{12} \{\ddot{\delta}\}_+ \quad 5.2.3.3$$

These may be used iteratively to estimate the future displacement and velocities. Initially, $\{\ddot{\delta}\}_+$ is estimated as $\{\ddot{\delta}\}_o$, but as the iteration proceeds, a better estimate of the future acceleration is obtained by applying the future displacement in the equation of motion (equation 5.2.1.1). Iteration continues until adequate convergence is obtained. The calculation is then stepped forward to the following time step. A simple first order interpolation polynomial is used for the first time-step when no '-' solution exists. This procedure is favoured because it enables boundary conditions to change in time (e.g. stiffness and inertia of a released node), simply by accommodating updated values of [M] and [K] in the equation of motion. Having obtained displacements and velocities throughout the specimen at each time-step, it is then possible to evaluate the total strain and kinetic energies whenever they are required.

Initially, the program reads in experimentally recorded crack length vs time data. These can be specified as times at which the crack path nodes should be released. At the instant of a node release, a force F_h (equation 5.2.2.1) equal and opposite to the force currently acting on that node is applied, and is decayed to zero as the crack grows to the next node. Every time a node is released, the energy absorbed by the crack is computed by both global energy balance and summation of the local work done by the holding back force. Both methods must agree if the solution is stable (Hodulak et al [80], Keegstra[77], Crouch[86]). Non-agreement indicates that energy is not being conserved. The stress intensity factor K^{dyn} can be calculated from the crack tip force at the instant of release (Keegstra[77]) or from equation 1.2.2.5.

Noticing that a large time-step caused an unstable solution, Keegstra[77] suggested the following stability criterion for the magnitude of the time-step:

$$\Delta t \leq \frac{d}{8 C_L} \quad 5.2.3.4$$

d - smallest size of an element in the structure. Crouch[86] assessed the stability of his numerical solution from the difference between G^{dyn} calculated by a global energy balance and by the work done by the holding back force. For the simulation of an SENT specimen loaded under constant displacement conditions, he reported an upper time-step size equal to 30% of the time needed for a stress wave to propagate between two closest nodes. A lower value of 20% was found to be more stable for a relatively small increase in computer time. In current work the time-step of $d/10C_L$ is used throughout.

5.2.4 Choice of elastic modulus

The materials involved in this study are strain rate dependent (table 2.2.1). Accurate numerical modeling of RCP in such materials would require a precise knowledge of the material behaviour at large strains and high strain rates. A number of authors have attempted to allow for viscoelasticity (or viscoplasticity) in their dynamic solutions. These are generally restricted to idealized cases, because the parameters of such models are usually unknown in real materials under RCP conditions (Crouch[86], Genussov[89]). For this reason, the majority of numerical solutions have concentrated on rate independent elastic calculations, i.e. a single constant modulus is assumed in the analysis. The question now arises of whether the static or the dynamic value of modulus should be used.

Popelar and Kanninen[80] showed that for a specimen loaded slowly, 'the long term modulus' (or static modulus - E_s) characterizes the conditions at the initiation of rapid crack growth, but 'the short term modulus' (or dynamic modulus - E_d) dominates during dynamic crack propagation. They noted that "during the unstable crack propagation event itself, viscous energy dissipation is quite small". Thus, elastic modelling of RCP in VE materials was justified.

The global response of a slowly preloaded and then suddenly broken sample can be interpreted as loading at a slow rate and unloading at a much higher rate. This is illustrated in figure 5.2.4.1.

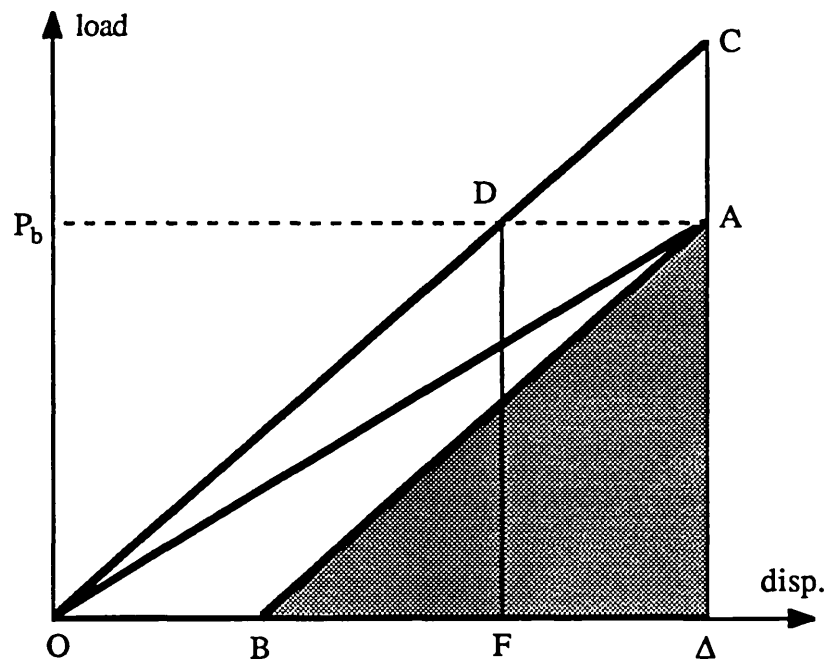


Figure 5.2.4.1: Schematic response of a VE material to a slow loading and fast unloading

The loading (line OA) is governed by a low modulus, while sudden unloading (line AB) is dominated by a higher modulus. The high modulus determines the amount of the retained available energy which can be released during the crack propagation (shaded area ABΔ). This

is only a fraction of the initially stored elastic energy (area OAA). The single modulus computation of G_{ID} in VE materials should consider only the available energy and not the total amount of energy stored in the sample prior to fracture. If samples are loaded slowly until RCP occurs (point A), and kept under constant displacement boundary condition during RCP, switching from the static modulus to the higher dynamic modulus at the instant of the crack initiation would create false stored strain energy (area OCA). For this reason, Crouch[86] used the static modulus only, although he recognized that the high rate modulus should be used for viscoelastic materials fracturing at high speeds. He based the calculation of G_{ID} on the single static modulus and fixed displacement boundary conditions. This single constant static modulus analysis generated low stress wave speeds and overestimated inertia effects. Furthermore, the static modulus calculation resulted in an available energy equal to the total stored energy (OAA). This was much larger than the true available energy (ABA), giving an upper bound value for G_{ID} . Genussov[89] challenged the correctness of the static modulus use in the modelling of RCP. He concluded that a better alternative is to incorporate a single dynamic modulus throughout the computation. The sample is loaded until the breaking force P_b is reached and unloaded along the same path (line BA). Although the applied displacement is only a fraction of the true displacement Δ , the total elastic energy (ODF) is similar to the true available energy. As in Genussov's numerical simulation, the dynamic modulus is chosen for the computations of RCP parameters in this work.

5.3 Elastic generation mode analysis of experimental results

5.3.1 General features

Crouch's version of Keegstra's program is used to analyse experimental RCP results of various specimen geometries (chapters 3 and 4). The program was modified to produce a version suited for the modelling of RCP in the duplex PMMA/PC and glass-PC SENT samples (sections 5.3.2, 5.3.4). Further modifications were necessary in order to: i) allow for the motion of the clamped boundary during RCP, i.e. to assess whether the loaded boundary is likely to remain fixed as the crack propagates (section 6.2.2), ii) model the sample with partially gripped boundaries (section 6.2.1), and iii) introduce a thin layer with small Young's modulus along the crack path, for the simulation of near crack tip non-linear behaviour of the materials (section 6.3.4).

The program is operated in the generation mode. The sample is assumed to be an ideally elastic material under plane strain conditions. Due to the symmetry, only half of the specimen is modelled. The specimen is represented by a FE mesh which is gradually refined towards the prospective crack path (figure 5.3.1.1). The elements of constant spacing are used along the crack path, i.e. a uniform mesh containing three rows of elements is used around the crack path line (figure 5.3.1.1). The mesh contained no abrupt changes in element size to avoid the generation of spurious stress wave reflections (Brickstad[83]). Since the smallest element governs the allowable time-step (section 5.2.3), the reduction in the element size is limited to a larger degree than in the static FE simulation, as computational cost would become excessive. However, the size of the elements along the crack path (the size of the smallest element was 0.11 mm), the grade of refinement and so the number of the elements (up to 5000 CST elements were used) were principally dictated by the specimen size. Care was taken to ensure that the dimensions of the real specimen were accurately represented.

The program has two modes of computation: static and dynamic. During the static loading, a uniform displacement is applied to the upper boundary nodes in the -y- direction (figure 5.3.1.1). The applied deformation is such that the resulting applied load is equal to the measured breaking load. During the static mode computation, the upper boundary nodes are constrained in the -x- direction. The nodes along the initial notch are not constrained, while all other crack path nodes (sample ligament) are constrained in the -y- direction, including the glass centre-line nodes in the case of the glass-PC SENT specimen analysis (section 5.3.4). After the static nodal displacements have been calculated by an iterative Gauss-Seidel procedure, the program switches to the dynamic mode by sudden release of the crack tip node. When the glass-PC SENT sample is modelled, the dynamic mode begins by sudden release of the glass centre-line node nearest to the crack mouth (section 5.3.4). The specimens are regarded as being under fixed-grip conditions during RCP. Therefore, in the dynamic mode computation, all the upper boundary nodes are kept constrained in both directions. The crack path nodes of the sample ligament are released successively according to the crack length history. G^{dyn} and crack speed are assumed to be constant over each nodal spacing along the crack path.

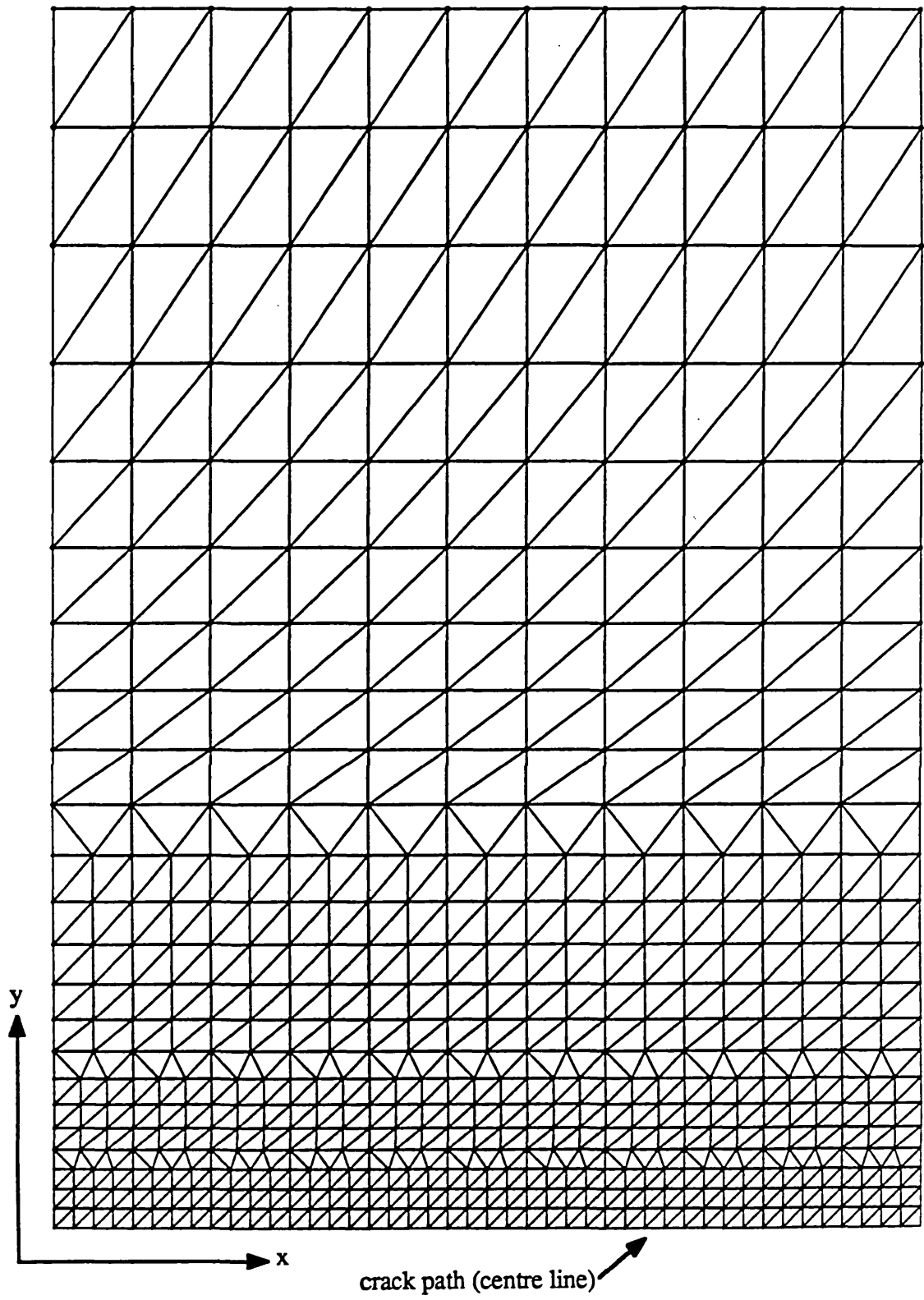


Figure 5.3.1.1: Finite Element mesh

5.3.2 Elastic analysis of duplex PMMA/PC SENT specimen

The duplex PMMA/PC SENT specimen is modelled by a FE mesh divided vertically in several columns (figure 5.3.1.1). The PMMA mechanical properties (E , ν , ρ) are specified for all elements on one side of the interface while all elements on the other side of the interface have the mechanical properties of the PC (table 2.2.1). The interface is represented as a row of nodal points among the elements with different E , ν , ρ . Thus, the interfacial bonding conditions are assumed to be ideal, i.e. no slippage is allowed between the PMMA and the PC.

Instants, at which the crack path nodes should be released, are specified in accordance with recorded crack length histories (section 3.5.2). The experimental records provide crack length vs time data beginning at a finite distance from the initial crack tip position. Due to the inability to monitor the details of the crack behaviour at the initiation, some assumption about the initial crack speed must be made. A number of authors assumed that the crack initiated at a speed equal to the first measured speed (Crouch[86], Wheel[91]). In order to investigate the sensitivity of calculated G_{ID} to the crack behaviour at initiation, three different crack initiation cases are considered. The crack is assumed to initiate at:

- a) a speed equal to the recorded speed - a constant speed crack initiation case (c.s.c.i.s.),
- b) zero speed followed by a gradual crack acceleration to the measured speed - an accelerating crack initiation case (a.c.i.c.),
- c) a high speed and gradually decelerate to the true crack speed - a decelerating crack initiation case (d.c.i.c.).

Figures 5.3.2.1, 5.3.2.3 and 5.3.2.5 represent the crack length histories of: i) the crack arrested at the interface, ii) the crack which propagated through the entire specimen and iii) the crack arrested in the PC part of the duplex sample (DS), respectively. Corresponding G_{ID} results are presented in figures 5.3.2.2, 5.3.2.4, 5.3.2.6. The details about geometry, breaking load and crack speed of the tests are given in tables 3.2.1 and 3.5.2.1.

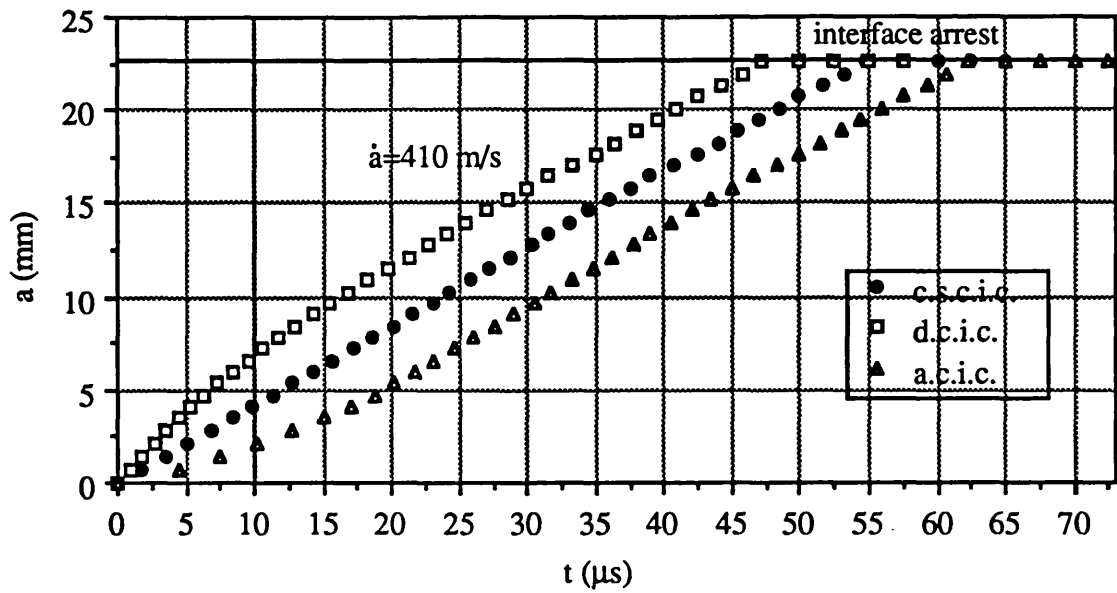


Figure 5.3.2.1: Crack length increment vs time; various initiation cases; test TBW6-9

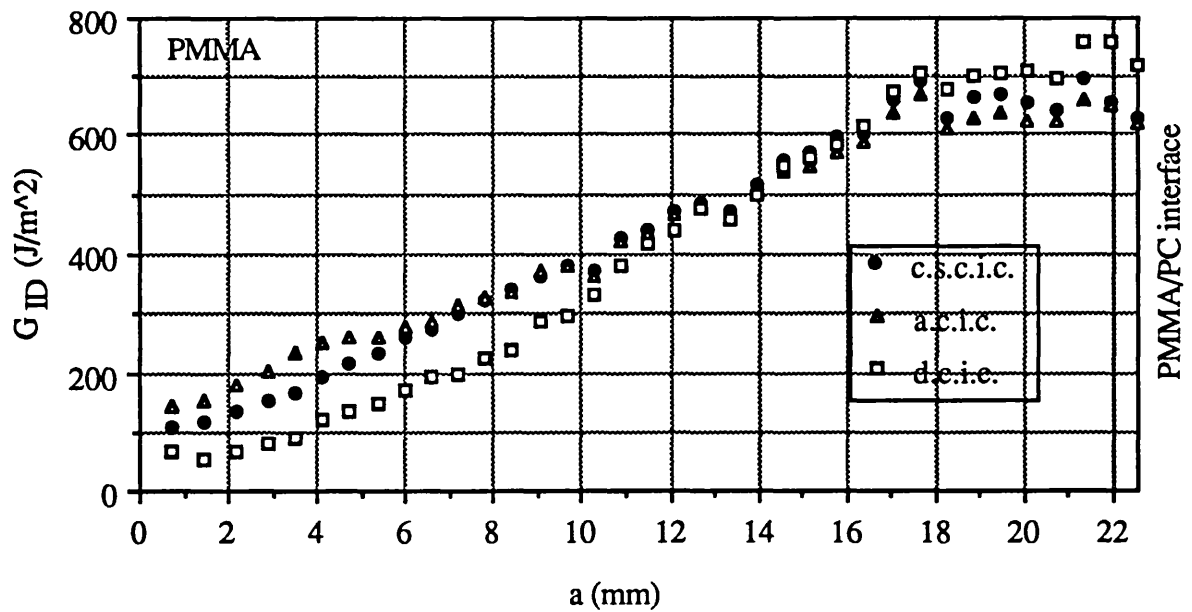


Figure 5.3.2.2: Crack resistance vs crack length increment; various initiation cases; test TBW6-9

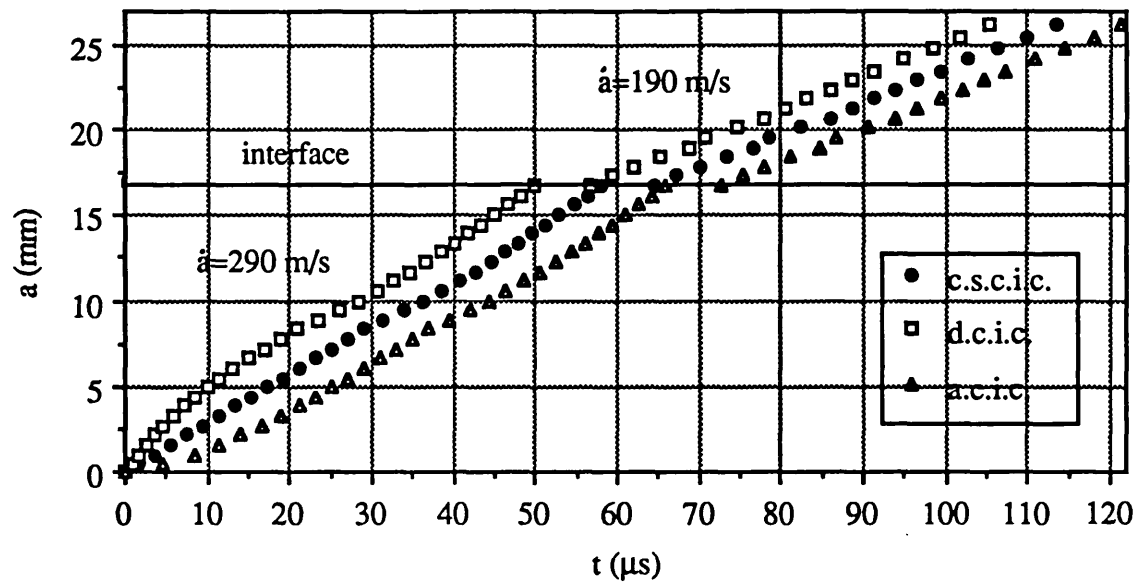


Figure 5.3.2.3: Crack length increment vs time; various initiation cases; test TMN10-4

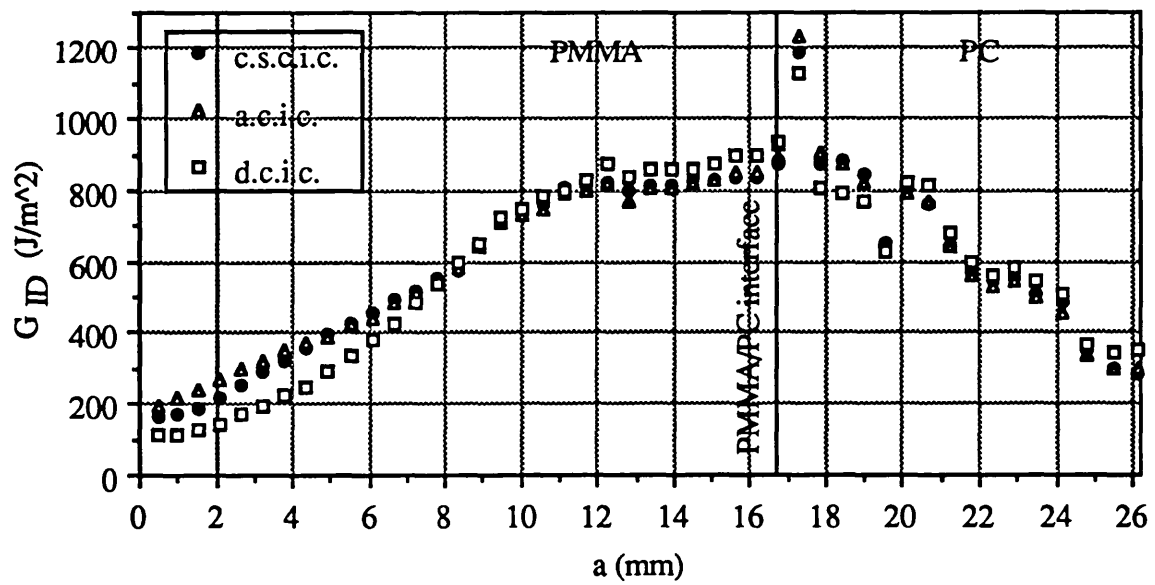


Figure 5.3.2.4: Crack resistance vs crack length increment; various initiation cases; test TMN10-4

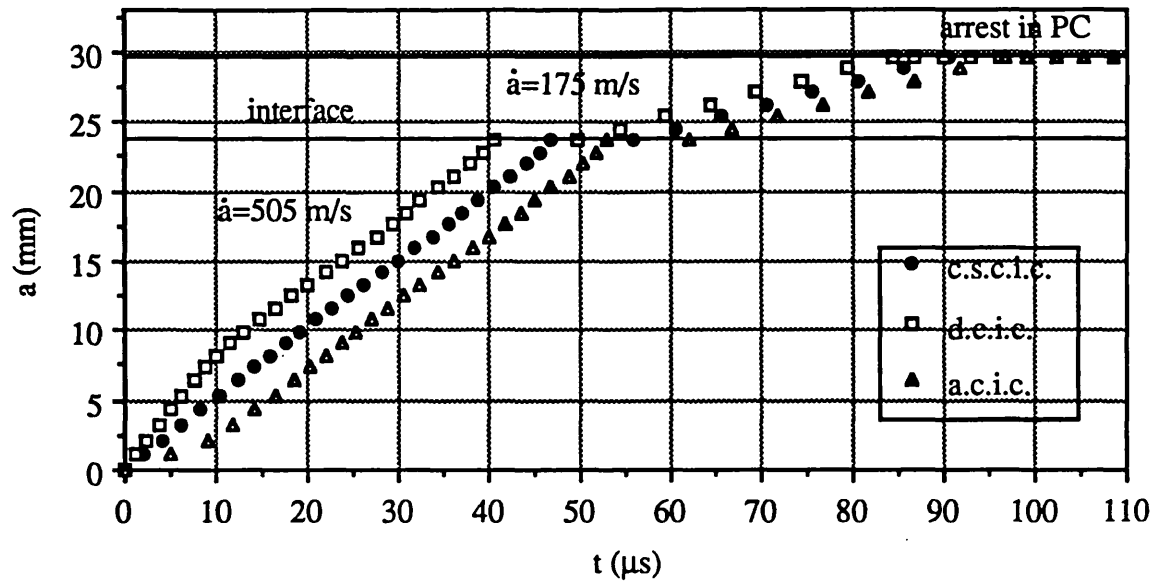


Figure 5.3.2.5: Crack length increment vs time; various initiation cases; test TBN10-6

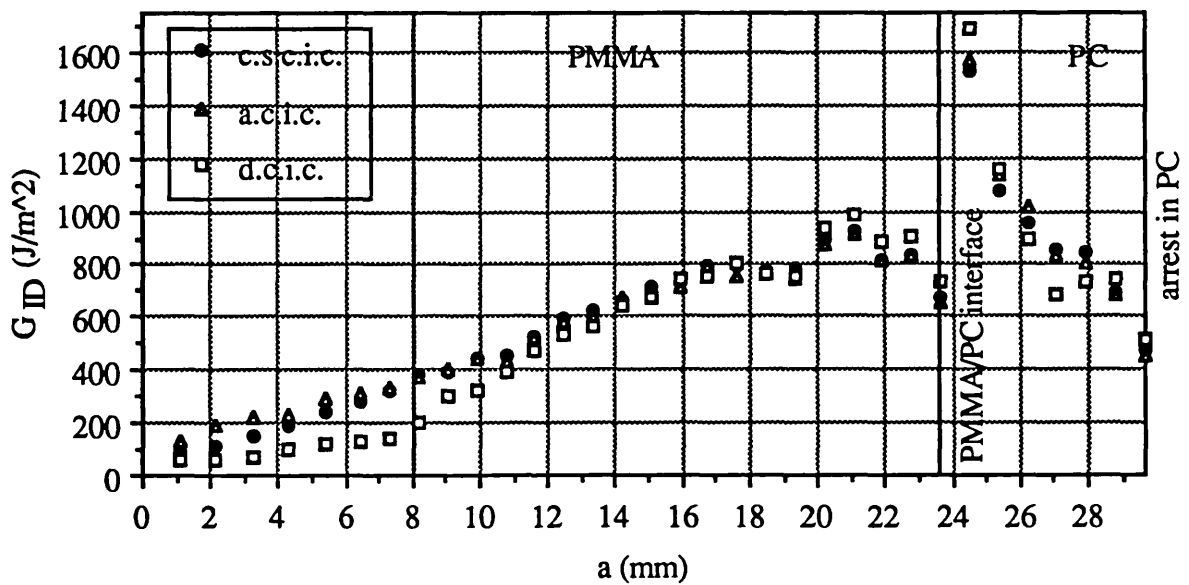


Figure 5.3.2.6: Crack resistance vs crack length increment; various initiation cases; test TBN10-6

The simulation of the crack tip node approaching the point of arrest is done by decaying the force F_h acting on the last released (adjacent) node to zero at the specified instant of arrest. The sequential nodal release process is terminated at the instant of arrest and the position of the crack tip node is fixed at that time. Similarly, the 'hesitation' of the crack at the interface is modelled by fixing the position of the crack tip node at the crack path interface node at time t_1 - the time when the crack tip reaches the PMMA/PC interface (figure 3.5.2.10). This node is kept constrained until time t_2 . At this instant, the node is released and the crack continues its propagation in the PC according to specified release times (t_2 is the time when the crack enters the PC, figure 3.5.2.10, section 3.5.2). So, during $t_2 - t_1$ the crack is simulated to be arrested at the interface.

Presented results (figures 5.3.2.2, 5.3.2.4, 5.3.2.6) showed that the various assumptions regarding the crack behaviour at the initiation stage (c.s.c.i.c., a.c.i.c., d.c.i.c.), resulted in initially different G_{ID} data, i.e. different crack speeds yielded different G_{ID} . Higher G_{ID} data corresponded to lower crack speeds (equation 1.2.2.3, section 1.2.2). As the assumed crack speeds approach the recorded crack speed, the differences between corresponding G_{ID} results decrease. Once the recorded speed is reached, subsequent G_{ID} values become rather insensitive to the assumed crack behaviour at the initiation stage. G_{ID} curves of all assumed cases (c.s.c.i.c., a.c.i.c., d.c.i.c.) follow the same trend. This indicates that the way the crack initiated is not critical in the analysis. Thus, in the rest of this work only the constant speed crack initiation case (c.s.c.i.c.) is used. Crouch[86] and Shmueli and Levy [82] also found a negligible effect of the initial crack acceleration, in SENT and DCB specimens, on G_{ID} behaviour. Given the speculative nature of the initial crack speed, it would be unwise to attach any significance to the calculated G_{ID} values near initiation.

The main characteristics of G_{ID} vs a curves of constant speed RCPs, presented in figures 5.3.2.2, 5.3.2.4, 5.3.2.6-7, are summarized as follows:

- a) Each figure demonstrates similar variation of G_{ID} with crack length in the PMMA part of the duplex sample. The calculated G_{ID} gradually rises (up to 8 times) as the crack advances. Eventually, after the irrotational stress wave, travelling at the speed of 2745 m/s in the PMMA, returns from the loaded boundary to the moving crack tip, G_{ID} stabilizes and remains almost constant. The numerical simulation of RCPs at higher crack speeds in low width to height ratio samples resulted in continuously increasing G_{ID} curves (figure 5.3.2.7).
- b) Figures 5.3.2.4, 5.3.2.6-7 show that immediately after the crack entered the PC part of the duplex specimen, the calculated G_{ID} has a high value. As the crack advances through the PC, G_{ID} decreases with the crack length, eventually stabilizes in the middle portion of the PC (figure 5.3.2.7), and then decreases further. It was shown that only G_{ID} values near the crack initiation, unlike the subsequent G_{ID} vs a behaviour, depend on the way the crack initiated. Thus, apart from the initial values, G_{ID} of the PC should be largely insensitive to the time of crack arrest at the PMMA/PC interface ($t_2 - t_1$). The influence of the duration of the crack 'hesitation' at the interface on G_{ID} vs a results of the PC part of the sample is shown in figure 5.3.2.7. It demonstrates the insensitivity of calculated G_{ID} data to an error in specified time ($t_2 - t_1$). The measured time ($t_2 - t_1$) was 3.4 μ s, and a value of 1.2 μ s was arbitrarily chosen (see section 6.3.7).

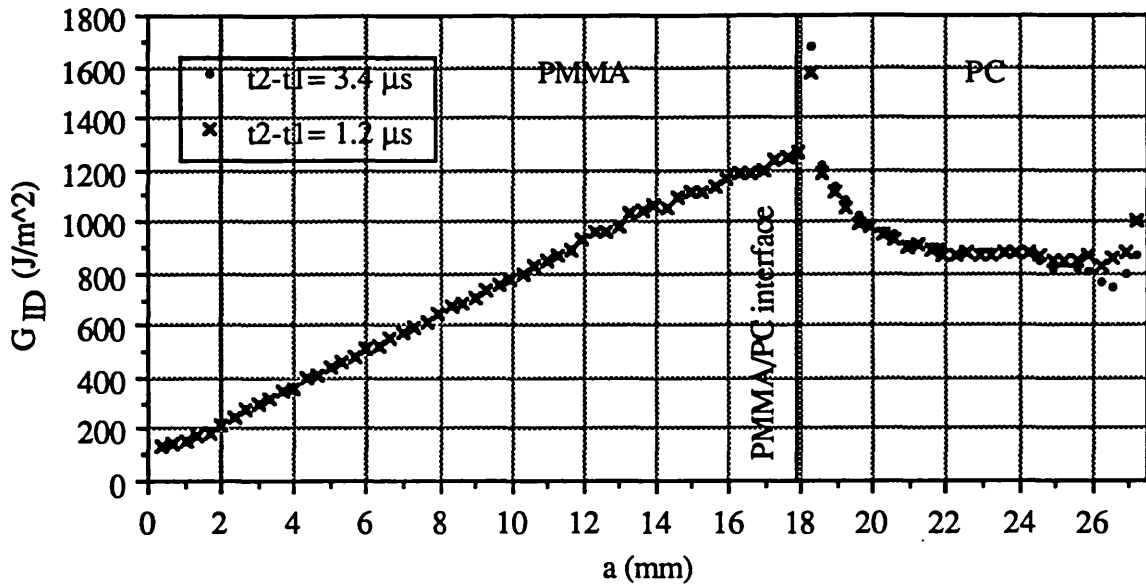


Figure 5.3.2.7: Crack resistance vs crack length increment; influence of the duration of the crack hesitation at the interface on G_{ID} of PC; test TMN10-2

The variations in the energy components, as obtained from the FE simulation of RCP in the duplex PMMA/PC SENT samples, are shown in figures 5.3.2.8 and 5.3.2.9. Figure 5.3.2.8 presents the results from the analysis of the completely fractured sample TMN6-3. The results from the modelling of test TMN6-5, where a permanent crack arrest occurred at the interface, are presented in figure 5.3.2.9. The main difference between these two cases is the amount of the initially stored elastic energy (see section 3.5.2). In test TMN6-5, the crack had much less available elastic energy, when it reached the interface, than in test TMN6-3.

Because of their inertia, the samples could not release all absorbed strain energy during RCP (test TMN6-3, figure 5.3.2.8).

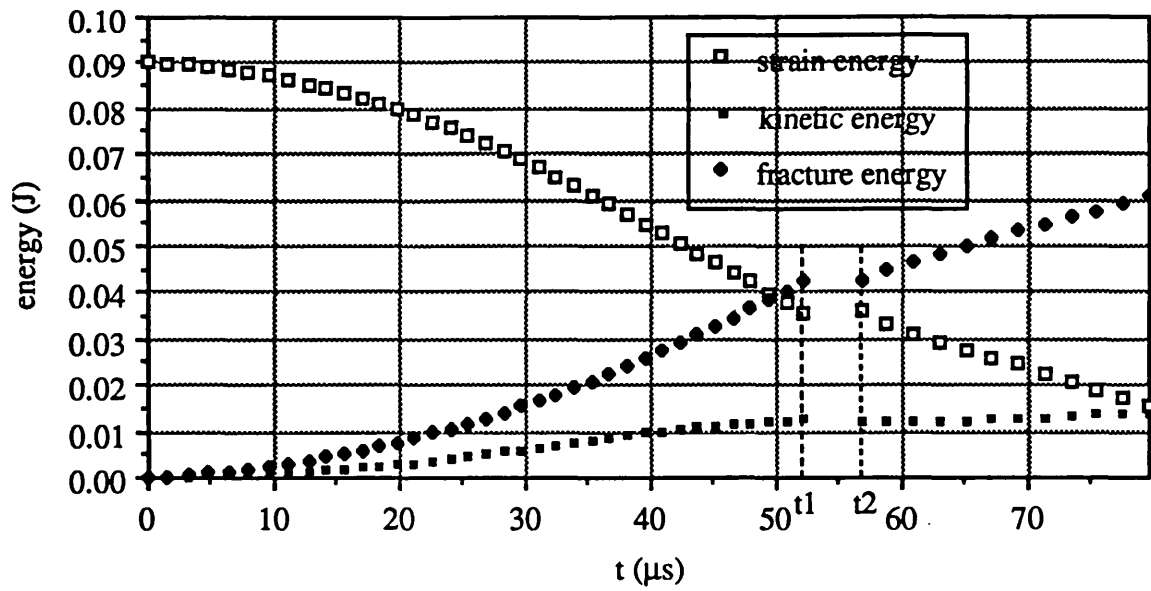


Figure 5.3.2.8: Energy vs time; test TMN6-3, (half specimen results)

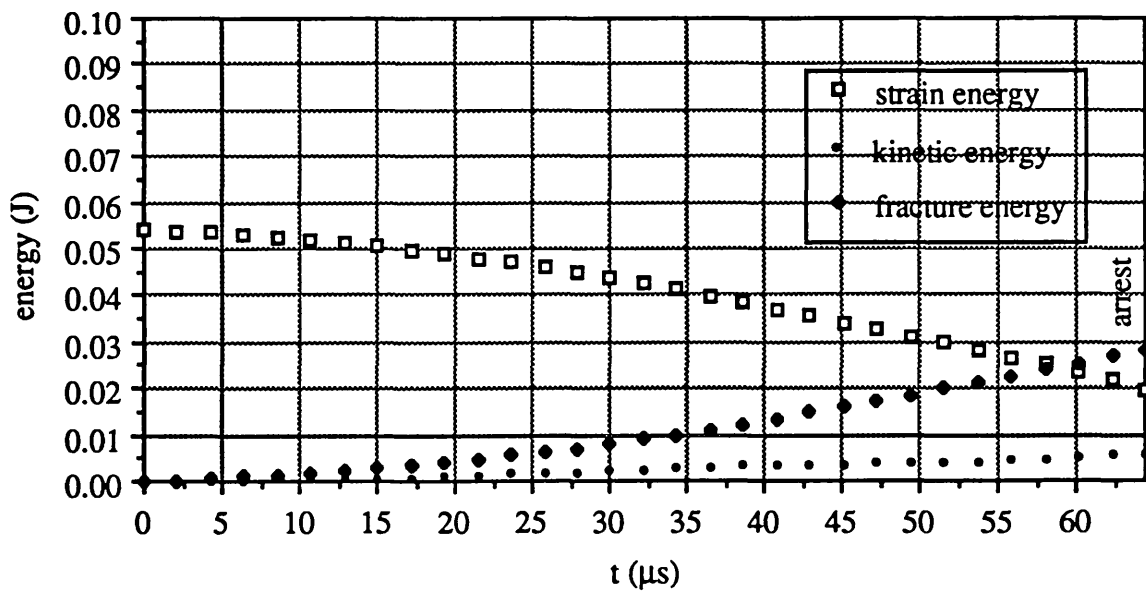


Figure 5.3.2.9: Energy vs time; test TMN6-5, (half specimen results)

5.3.3 Elastic analysis of PMMA SENT specimen

The grid line technique was used for recording the crack length histories in single material SENT specimens (sections 2.3.2, 4.3). Due to the nature of this technique, there are two factors which may impair the accuracy of the crack length data used in the FE simulation, and hence, influenced numerical G_{ID} results: i) non-deterministic initiation event, and ii) the incremental character of the crack history reading (timing lines were spaced every 1 - 2 mm, section 4.3). The effect of various types of crack behaviour at initiation on subsequent behaviour of G_{ID} was demonstrated in the previous section. In order to examine the influence of the second factor on G_{ID} results, two different interpretations of the crack length record are considered. Firstly, the nodal release times are evaluated assuming a constant crack speed between the measured data points - discrete data (figure 5.3.3.1). In the second computation, a straight line is fitted to the experimental points (figure 5.3.3.1). The corresponding G_{ID} results are presented in figure 5.3.3.2. In both simulations, the initiation time is obtained using c.s.c.i.c. extrapolation (see previous section). As shown in figure 5.3.3.2, the generated G_{ID} values are in reasonably good agreement - the difference is within the range of $\pm 10\%$. So, bearing in mind the large variation in G_{ID} with crack length (up to 800%), this uncertainty is almost negligible. Thus, in the rest of this work, the linear approximation of incremental crack length records is used due to its simplicity. In further support to this decision, it should be emphasized that experimental crack length vs time points fell very close to a straight line in most of the tests (section 4.4.1), as demonstrated in figures 5.3.3.3 and 5.3.3.4.

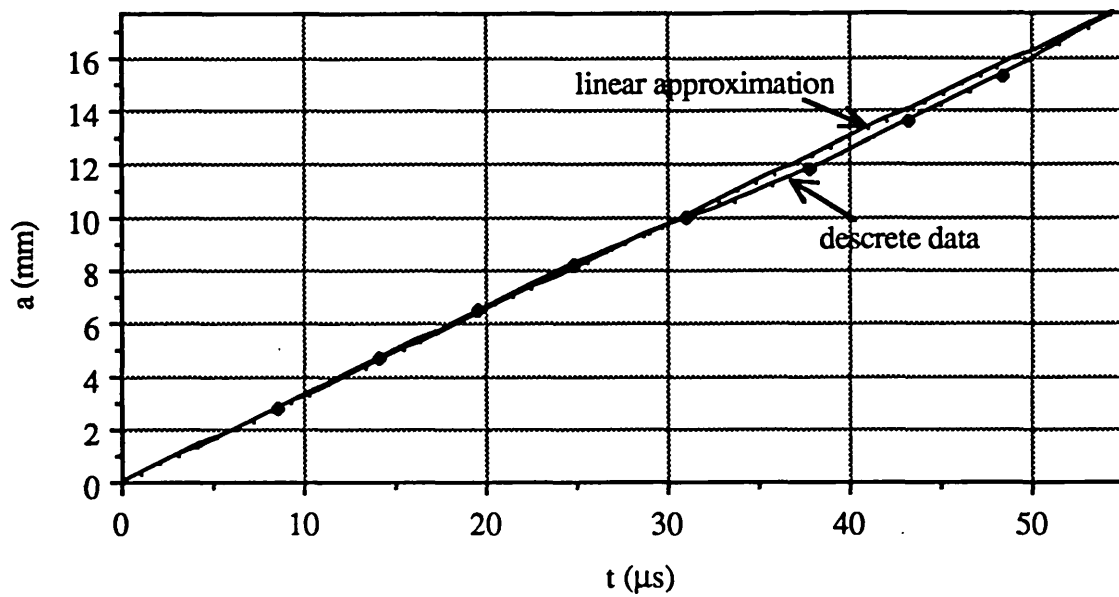


Figure 5.3.3.1: Crack length increment vs time; various interpretations of the experimental data

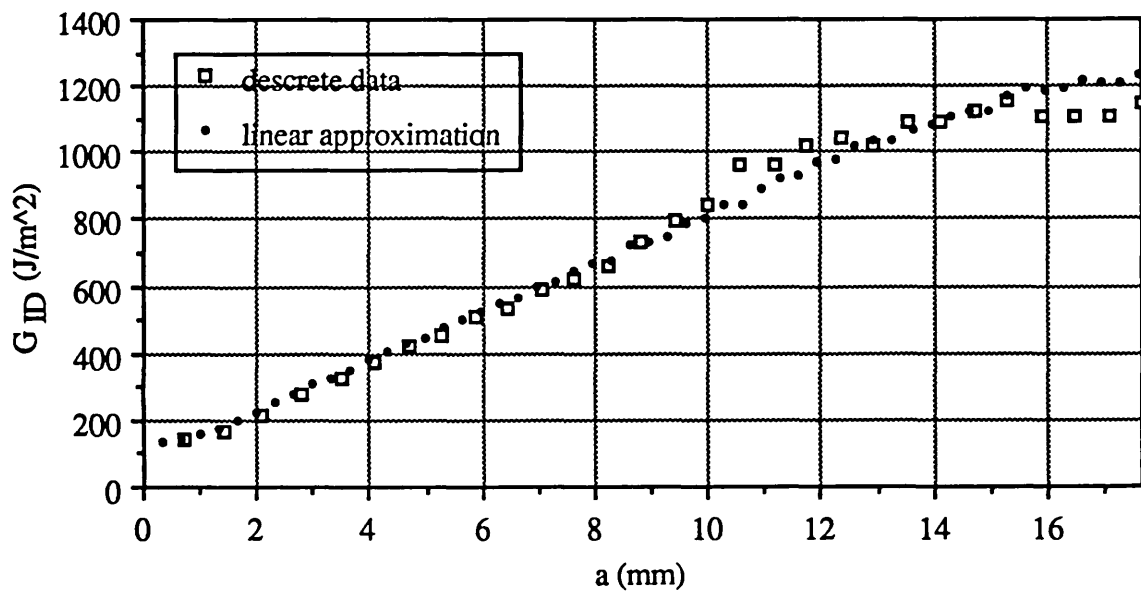


Figure 5.3.3.2: Crack resistance vs crack length increment; influence of the data interpretations

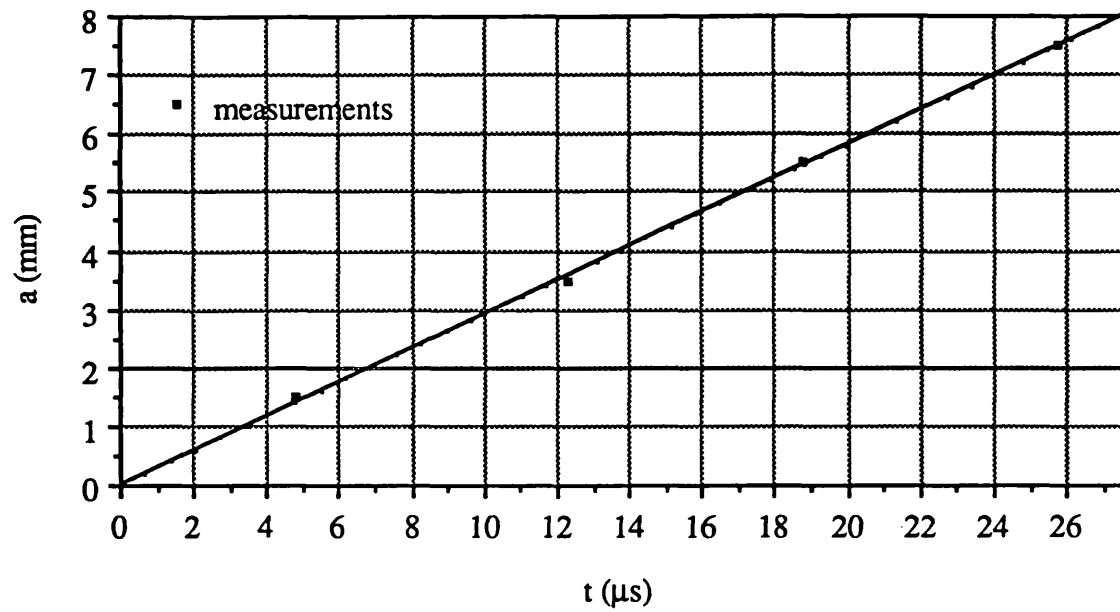


Figure 5.3.3.3: Crack length increment vs time; test PM4-4

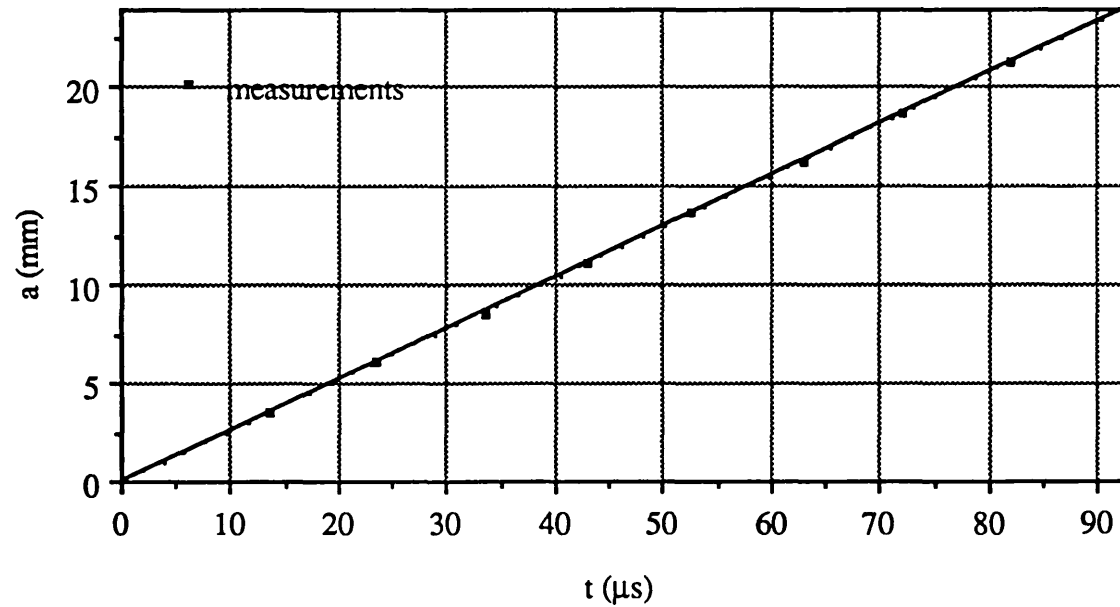


Figure 5.3.3.4: Crack length increment vs time; test PM5-1

In order to analyse the influence of the specimen size on G_{ID} results, a number of fixed displacement tests on the PMMA SENT specimens of various D/H ratios were performed (sections 4.2.1, 4.4.1). These are analysed numerically, and selected results are presented in figures 5.3.3.5 and 5.3.3.6. The most marked feature of the generated G_{ID} curves is that each curve has a rising trend over a significant portion of the crack growth. This rising part is eventually followed by a G_{ID} plateau region. The length of the rising portion of G_{ID} curve, is found to be longer for specimens with lower D/H ratios and higher crack speeds. This was observed in the simulations of the PMMA/PC SENT specimens as well (section 5.3.3.2). Another significant feature of the results (figures 5.3.3.5, 5.3.3.6) is that a greater G_{ID} variation at a higher average value occurs in the taller specimens (lower D/H ratio) at the same crack speed. Crouch[86] reported similar results from numerical tests on the influence of the PMMA SENT specimen height on crack resistance. He also used the infinite plate expression to analyse the tests (equation 1.2.2.3), and reported good agreement with numerical results even after return of the stress wave from the loaded boundary to the crack tip. More recently, Hren[90] observed the same influence of the specimen height on G_{ID} of an epoxy resin, as reported above. He achieved an extremely large variation in the height of the specimens by using epoxy-aluminium adhesive joints with glue line thickness in the range of 0.5 to 5 mm. He then switched to epoxy SENT configuration and tested samples with a height up to 100 mm.

Figure 5.3.3.6 shows the comparison between numerical and analytical results. Specimen PM8-3 is analysed using the dynamic infinite plate solution (equation 1.2.2.3) - line 1, and the static infinite plate expression (equations 1.2.1.6 and 1.2.1.2) - line 2. In the former case, G^{st} is defined by equation 1.2.1.6, while the crack speed correction factor $g(\dot{a}) = 0.88$ is given by Broberg[67]). The dynamic infinite plate solution should be valid until a stress wave reflected from a boundary returns to the tip. Waves reflected from loaded boundaries are likely to have a

greater influence than waves reflected from free edges (Crouch[86], Genussov[89]). Values from the dynamic infinite plate prediction are in close agreement with the results from the numerical simulation of the finite geometry. Until the stress waves return to the crack tip, the crack is 'unaware' of the specimen size. As the stress waves returned to the tip, some divergence starts to develop, although no abrupt step change due to the wave arrival itself occurs.

The infinite strip solution (Nilsson[72], Bergkvist[73]), involving no dynamic correction, might be applied to a tensile specimen of high D/H ratio, with longitudinal contraction prevented due to rigid loading clamps. For a specimen of height H loaded to a uniform displacement Δ along its length under plain strain conditions, G is given as:

$$G = \frac{\Delta^2 E (1 - \nu)}{2 H (1 + \nu) (1 - 2\nu)} \quad 5.3.3.1$$

Also shown in figure 5.3.3.6, is a comparison between the numerical results of test PM5-1 (D/H=0.75) and the infinite strip approximation - line 3. The numerical solution approaches G given by equation 5.3.3.1, and is below the true strip case. From a similar numerical study of SENT specimens of various D/H ratios, Dahlberg et al. [80]) concluded that dynamic effects could only be ignored in specimens of high D/H ratio.

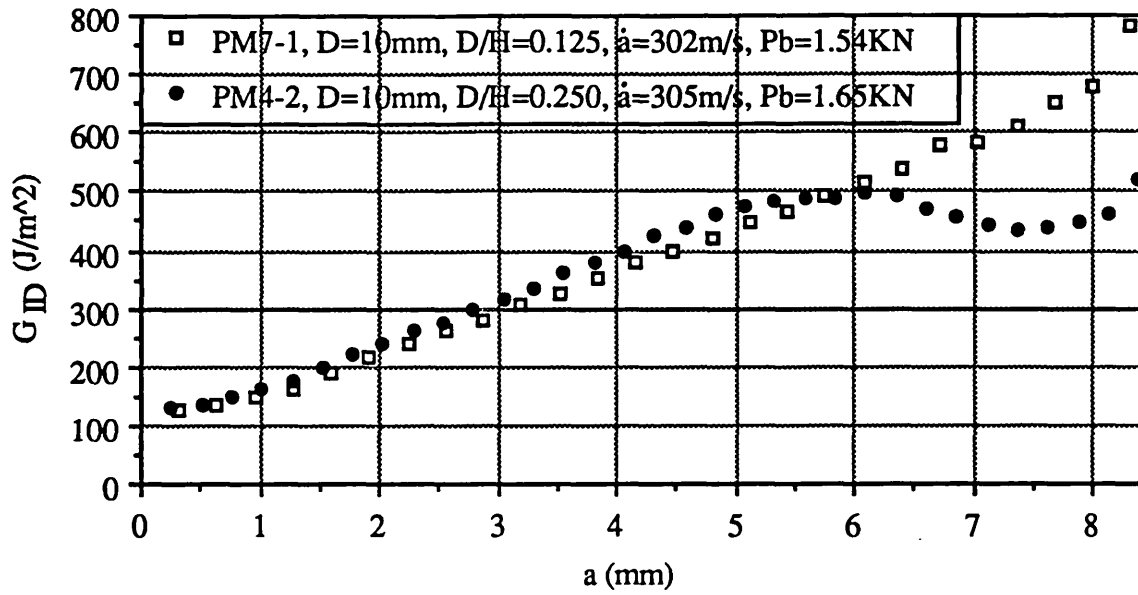


Figure 5.3.3.5: Crack resistance vs crack length increment, influence of various D/H ratios

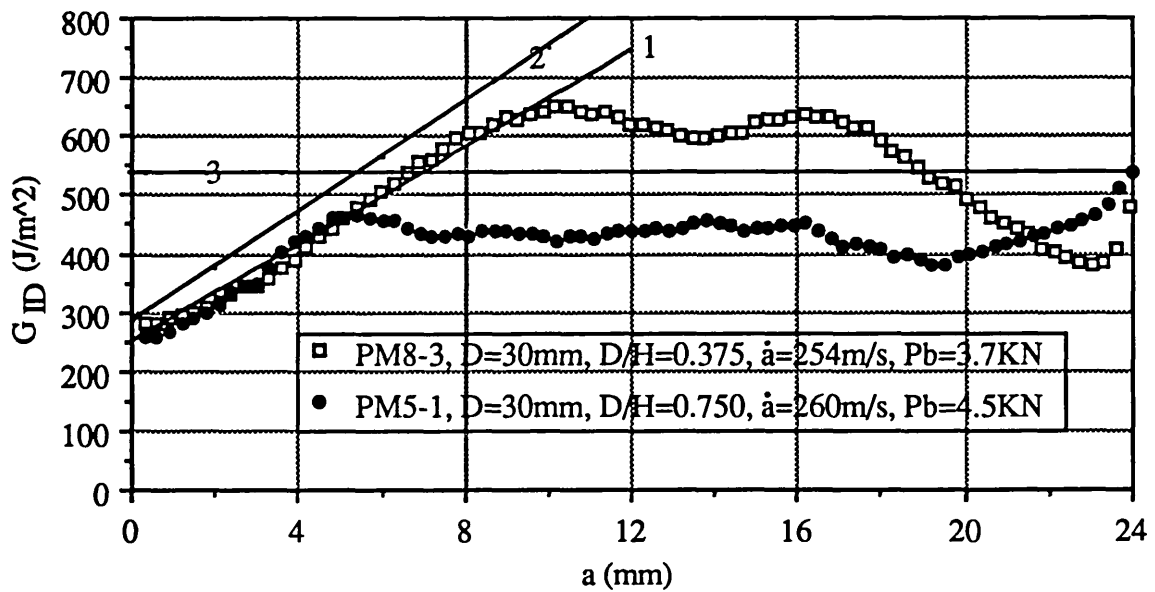


Figure 5.3.3.6: Crack resistance vs crack length increment, influence of various D/H ratios,

line 1 - the dynamic infinite plate solution (equation 1.2.2.3),

line 2 - the static infinite plate solution (equation 1.2.1.6),

line 3 - the infinite strip solution (equation 5.3.3.1).

Figure 5.3.3.7 demonstrates the insensitivity of calculated G_{ID} to the mesh size. Test PM4-2 is simulated by using two different FE meshes:

mesh 1 - consists of 1030 elements, the size of the smallest element (around the crack path) is 0.254 mm,

mesh 2 - consists of 425 elements, the size of the smallest element is 0.5225 mm.

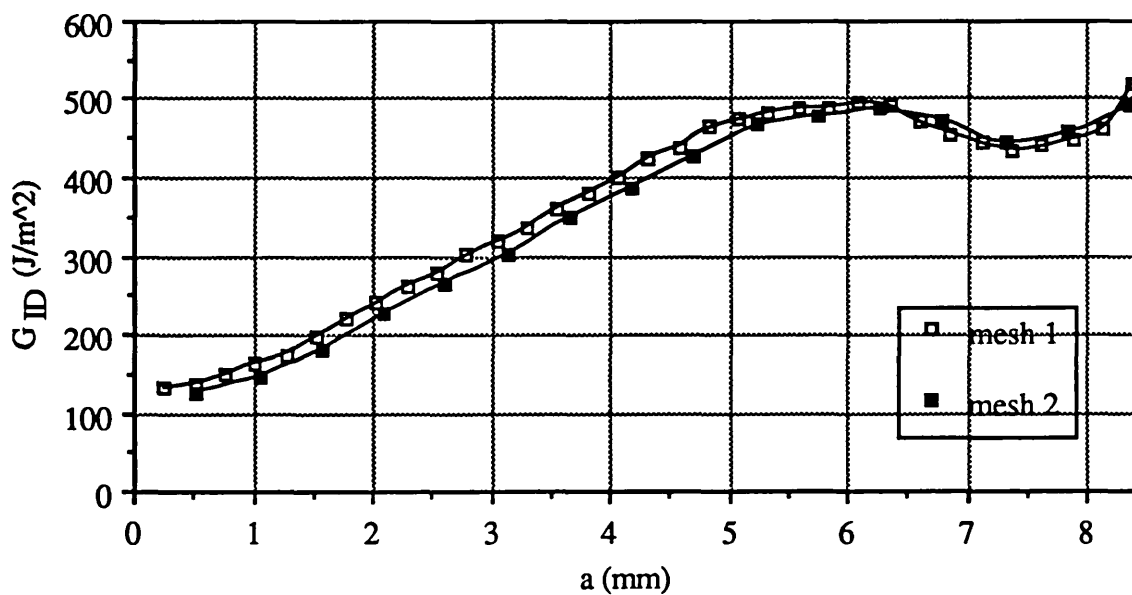


Figure 5.3.3.7: Crack resistance vs crack length increment, influence of the mesh size

5.3.4 Elastic analysis of glass-PC SENT specimen

Due to symmetry, only half of the glass-PC SENT specimen is modelled, and the PC part is represented by a FE mesh similar to the mesh shown in figure 5.3.1.1. The glass slide is modelled by additional elements of different mechanical properties; $E = 60 \text{ GPa}$, $\nu = 0.25$ and $\rho = 2500 \text{ kg/m}^3$, attached to the 'PC' mesh. Bonding conditions between the glass slide and the PC sample are assumed to be ideal (see section 5.3.1).

During the initial static computation which corresponds to slow loading of the sample to the prescribed breaking load, the crack path nodes of the PC ligament, including the glass centre-line nodes, were constrained in the direction of the load application. The nodes along the initial notch were not constrained. According to the observations reviewed in section 4.2.2, it is assumed that the crack initiates in the glass slide at the inner surface opposite the prenotch (point O, figure 5.3.4.1). This is the most highly stressed section of the glass. The program switches to the dynamic mode by sudden release of the glass centre-line node nearest to the crack mouth. A reference, zero time ($t_0 = 0$.) corresponds to the instant of the crack initiation in the glass slide, i.e. to the beginning of the dynamic mode calculation. The rest of the glass centre-line nodes are sequentially released according to assumed crack speed in the glass ($\dot{a}_{\text{glass}}$) of 1500 m/s (Genussov[89]). The energy absorbed by the crack propagation in the glass is neglected in the calculation. As the triggering line was placed on the outer surface of the glass, the time the glass completely fractured - $t_A = 2 \mu\text{s}$ (point A, figure 5.3.4.1), should coincide with the instant when the recording of the crack length history started. The time interval (T_{AC}) between the moment of triggering to the breaking of the first crack velocity gauge (point C, figure 5.3.4.1) is experimentally recorded in the range of 14 to $30 \mu\text{s}$ (section 4.4.2).

It was demonstrated in section 5.3.1, that the influence of various types of crack behaviour at

the initiation on subsequent behaviour of G_{ID} is insignificant. The experiments on the glass-PC SENT geometry showed that the crack advanced at a remarkably constant speed (section 4.4.2). The effect of different interpretations of the incremental crack history reading on G_{ID} is assessed in the previous section.

Taking the above arguments into account, the nodal release times of the PC crack path nodes are evaluated by fitting a straight line through the experimental points. Thus, the PC initial crack tip node (point B, figure 5.3.4.1) is released at a time t_B , given by:

$$t_B = t_A + T_{AC} - \frac{a_{BC}}{\dot{a}_{PC}} \quad 5.3.4.1$$

where: a_{BC} is the distance between the initial crack tip (point B) and the end of the first crack velocity gauge (point C), and $\dot{a}_{PC}$ is the crack speed in the PC.

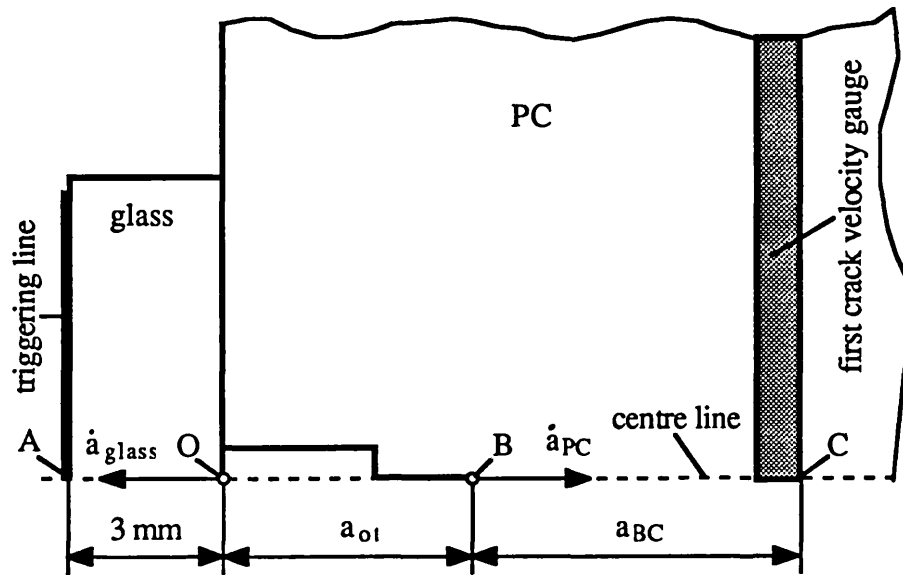


Figure 5.3.4.1: Glass-PC specimen and the crack initiation

Although Genussov[89] claimed that the breaking of the glass could take up to 10 μ s, he instantly released the glass centre-line nodes, setting the crack velocity in the glass to infinity. This simplified the simulation of the glass-PE specimen response. In so doing, possible effects

of the true non-instantaneous breaking of the glass, on computed G_{ID} values of the PE, were neglected. An assessment of the influence of the crack behaviour in the glass slide on the crack resistance results is therefore desirable. Test PC4-5, with $T_{AC} = 14 \mu s$, $a_{BC} = 1.4 \text{ mm}$ and $\dot{a}_{PC} = 406.7 \text{ m/s}$, is numerically simulated assuming that:

test 1 - the glass-slide instantly breaks, thus $t_0 = t_A = 0$ and $t_B = 10.56 \mu s$,

test 2 - the crack initiated at the outer surface of the glass (point A) and propagates at the speed of 1500 m/s , thus $t_A = 0$, $t_0 = 2 \mu s$ and $t_B = 10.56 \mu s$,

test 3 - the crack initiated at the inner glass slide surface (point O) and advances at the speed of 1500 m/s , thus $t_0 = 0$, $t_A = 2 \mu s$ and $t_B = 12.56 \mu s$.

The corresponding results are presented in figure 5.3.4.2.

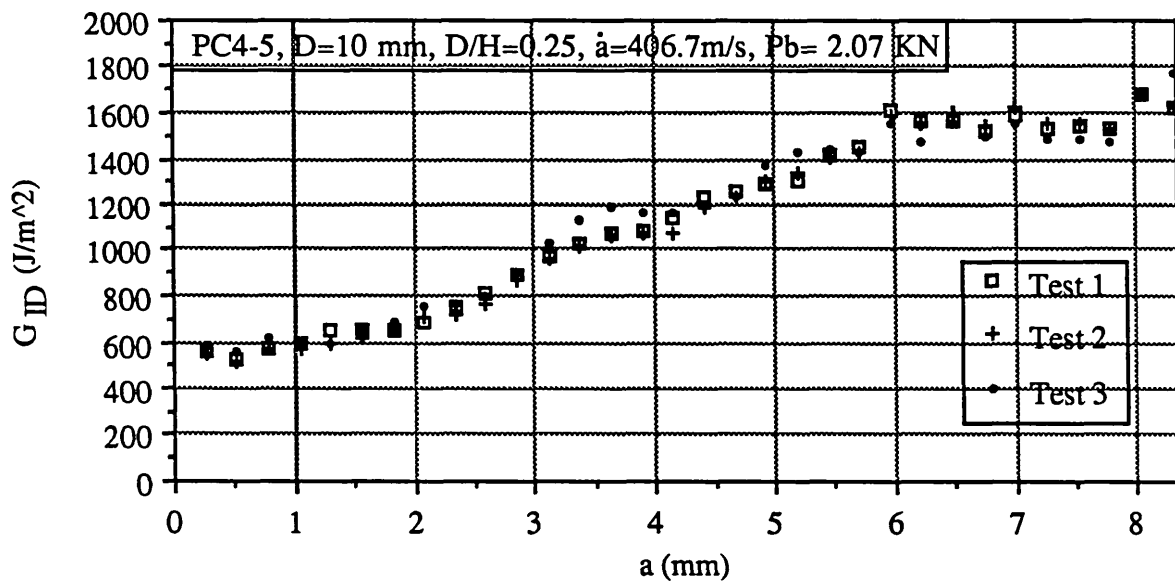


Figure 5.3.4.2: Crack resistance vs crack length increment, influence of the crack behaviour in the glass-slide on G_{ID} of PC

The remaining factor which may influence the initiation time t_B (equation 5.3.4.1), is the time

delay (T_{AC}) between the moment of triggering to the first step change of voltage in the staircase like voltage vs time record (this first step corresponds to the breaking of the first crack velocity gauge). The accuracy of the T_{AC} value is governed by the quality of the resistance reading (section 4.4.2). The effect of erroneous T_{AC} is investigated by evaluating G_{ID} of the sample PC4-5, for three different values of T_{AC} . These are the measured value of $14 \mu s$ and assumed values of $14 \pm 5 \mu s$ (figure 5.3.4.3).

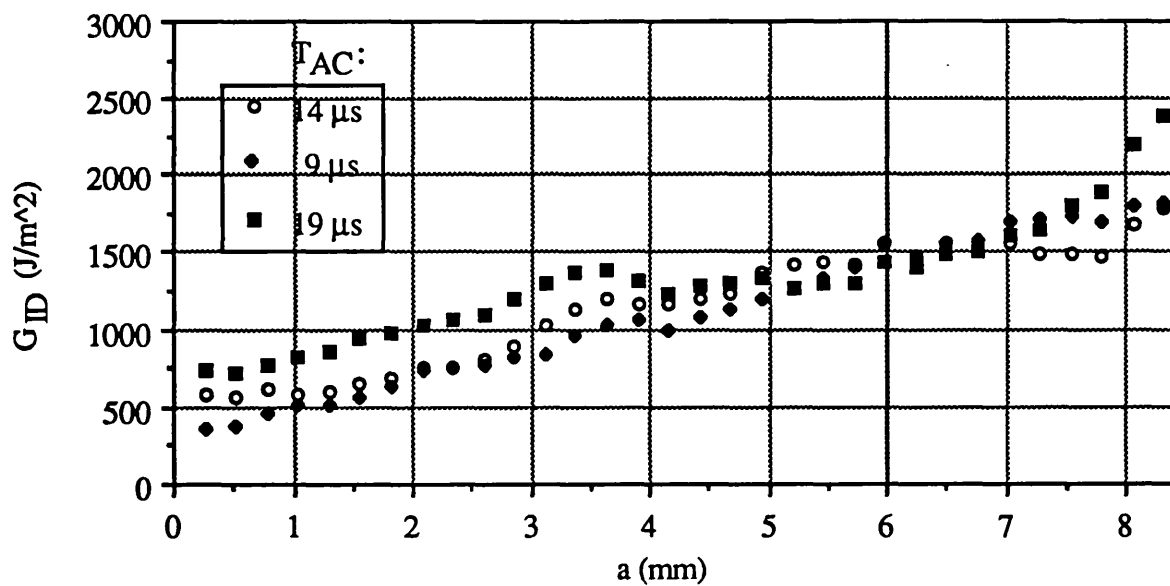


Figure 5.3.4.3: Crack resistance vs crack length increment, influence of T_{AC} on G_{ID} of PC

Figures 5.3.4.2 and 5.3.4.3 demonstrate that the generated G_{ID} vs a curves are rather insensitive to both the behaviour of the crack in the glass slide and the error involved in the measurement of the T_{AC} . This is shown to be true even when a high level of uncertainty is involved in defining the above two parameters.

Selected G_{ID} vs a results, from the numerical simulation of RCP in the glass-PC SENT specimens, are summarized in figures 5.3.4.4-5. The main features of the G_{ID} vs a curves obtained from the PMMA samples and the PMMA part of the PMMA/PC specimens, are also observed in the analysis of glass-PC specimens. These features were explained in detail in sections 5.3.2 and 5.3.3, and are not repeated here.

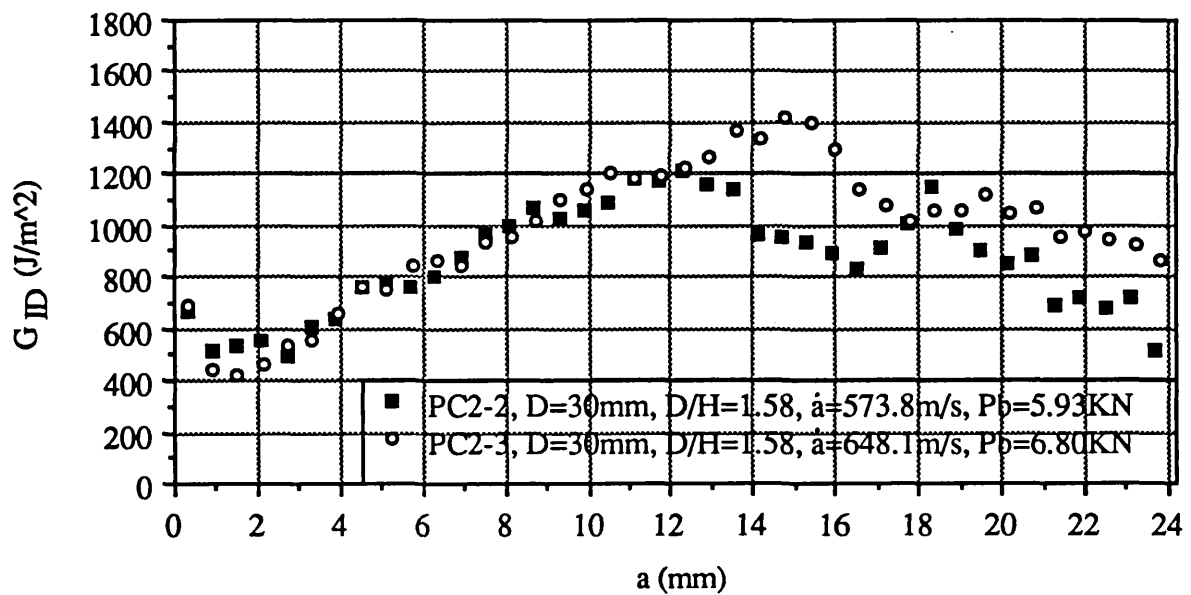


Figure 5.3.4.4: Crack resistance vs crack length increment, influence of the crack speed

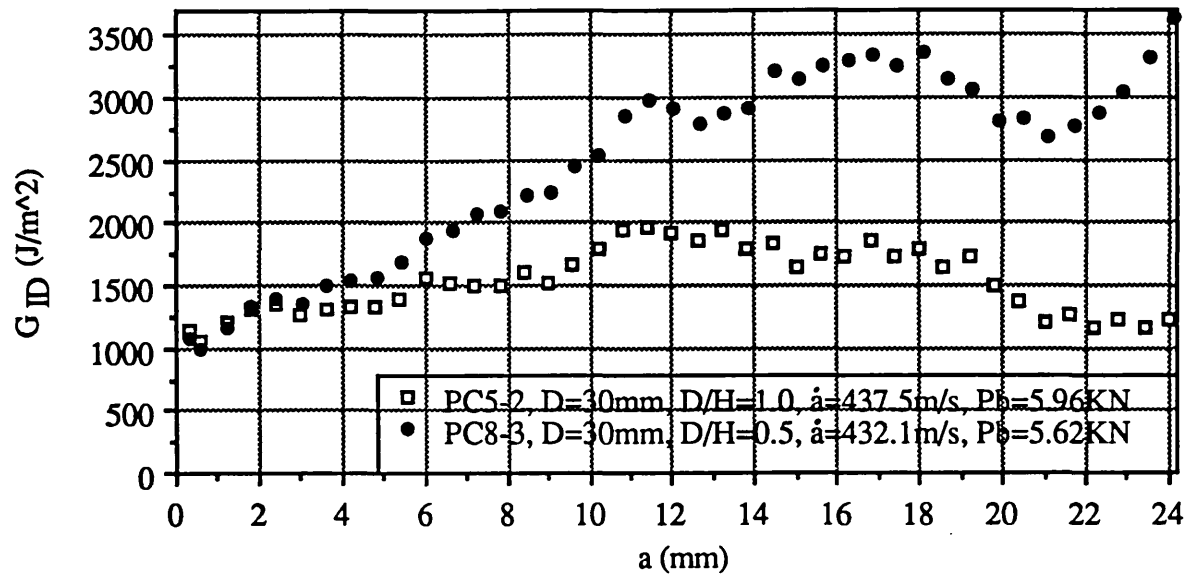


Figure 5.3.4.5: Crack resistance vs crack length increment, influence of the specimen size

For crack speeds in the range of 350 to 420 m/s in PC SENT samples, Crouch[86] reported the lower bound G_{ID} value of about 800 J/m². Kobayashi et al [80] obtained the value of about 960J/m² for crack speeds up to 400 m/s. They used compact tension PC specimens.

CHAPTER 6

'LOCAL MODULUS' MODEL

6.1 Introduction

The crack length histories of the constant speed RCPs in the duplex PMMA/PC, the PMMA and the glass-PC SENT specimens (chapters 3 and 4), were analysed using the dynamic FE code employing a single modulus (chapter 5). The very large variations of G_{ID} with crack length and speed which are implied seem to be particularly problematical. Generated results cast doubt on the uniqueness of the computed crack resistance. This outcome of the elastic analysis of the experimental results is thought to be unacceptable since rather uniform fracture surface appearances are associated with constant speed RCPs.

Chapter 6 demonstrates that the major error in the analysis employing a single constant modulus is inadequate representation of the material behaviour around the moving crack tip. It describes an attempt to overcome these difficulties by exploring the concept that the anomalies arise from large local strains around the crack tip resulting in the crack 'seeing' a low local modulus (section 1.3). This is done by using a thin local strip along the prospective crack path in the FE analysis. This has been discussed before, though not quantified in this way.

6.2 Discussion of the elastic analysis results

6.2.1 The nature of G_{ID}

Numerical G_{ID} results of the experiments on the duplex PMMA/PC, the PMMA and the glass-PC SENT specimens, were presented in sections 5.3.2, 5.3.3 and 5.3.4 respectively. These are rather confusing since the large variation of G_{ID} with the crack length and speed made it difficult to establish the intrinsic value of G_{ID} . Consequently, the condition for the crack propagation across the PMMA/PC interface of the duplex sample (equation 2.4.2.2), cannot be verified.

A tendency of the crack to propagate at a constant speed in various SENT configurations was noted in sections 3.5.2, 4.4.1 and 4.4.2. This is despite the very large variations of the crack driving force predicted across the specimen. This implies that during RCP the amount of energy fed into the crack may increase several times while the crack still propagates at a constant speed. On the other hand, the uniformity of the fracture surface appearance across the sample fractured at a constant speed (sections 3.5.2, 4.4.1, 4.4.2), indicates that a similar amount of energy was absorbed along the crack path (rather constant G_{ID} with crack length). Freund and Lee [90] speculated that the crack should accelerate if the applied K_{dyn} increases as the crack propagates. Ravi - Chandar and Knauss [84-b] observed crack speed constancy as G_{dyn} varied across the sample, for a broad range of crack speeds in Homalite and PMMA. Crouch[86] suggested that the crack probably reached its 'terminal' speed for the certain geometry and the variations in the crack resistance at constant speed could be interpreted as the vertical stem in G_{ID} vs a curve (due to adiabatic heating). This suggestion can easily be disproved since the range of crack speeds (200 to 400 m/s in the PMMA, 400 to 600 m/s in the PC), associated with wide G_{ID} variations, cannot be regarded as the terminal crack speed. As a

similarity was observed between the fracture surfaces of samples broken at the terminal crack speed but under different, predicted G_{ID} , a non-unique crack resistance vs surface roughness dependence was offered as the explanation for the vertical stem in G_{ID} vs $\dot{a}$ curve (Crouch[86]). However, this non-uniqueness was agreed to be somewhat unexpected. In contrast to this, Crouch[86] further reported that in a limited number of SENT samples, the crack speed was constant, calculated G_{ID} was approximately independent of the crack length, while surface roughness increased along the sample. This implies that G_{ID} correlates with the crack speed, while neither G_{ID} nor the crack speed correlate with the surface roughness. Bradley and Kobayashi [71] claimed that the surface roughness correlated with stress intensity factor rather than velocity. They reported a lag of crack speed changes behind changes in K_{dyn} occurred (this was a marginal effect). They further concluded that: "Since the stress intensity factor is a measure of the local load or strain energy, the change in surface roughness can be expected to correspond to the variation in the stress intensity factor.". Thus, the samples which fractured at similar crack speeds and experienced similar fracture surfaces should have absorbed similar amounts of fracture energy (similar G_{ID}). However, if different crack speeds are associated with different mechanical properties of tested material in the local zone (due to different applied strains, strain rates, temperature around the crack tip), it would be difficult to correlate the changes in the fracture surface roughness with changes in G_{ID} . As a consequence of the multivalued G_{ID} with crack length and speed, it would be more appropriate to present data as G_{ID} bands at each speed rather than as a single curve (Rosenfield and Kanninen [73], Ravi-Chandar and Knauss [84-b, 85]).

Another problematical feature of the elastic results of chapter 5 is the G_{ID} dependency on the specimen size. It was demonstrated in sections 5.3.3 and 5.3.4 that at similar crack speeds, a greater G_{ID} variation at a higher average value occurred in taller specimens. In the numerical simulation, each specimen was regarded as being under fixed-grip conditions during RCP. After the crack is initiated, the highly tensile stressed region near the crack tip suddenly relaxes

and a compressive stress wave is generated. As the crack advances, compressive stress waves are continuously radiated from the wake of the moving crack. Until a reflected stress wave returns from the loaded boundary to the crack tip, the crack is 'unaware' of the specimen height and loading conditions and the failure is effectively at a constant applied load. Consequently, G^{dyn} increases with the crack length. As the stress waves reach the crack tip, the failure is governed by constant displacement conditions, and a stabilization of G^{dyn} with crack length begins. Thus, in the case of two specimens with different heights but with all other dimensions nearly the same, G^{dyn} stabilizes at a longer crack length and hence at a higher value in the taller specimen (sections 5.3.3-4). Figures 6.2.1.1 and 6.2.1.2 compare the variations of the energy components in short and tall PMMA specimens respectively. Kinetic energy and retained strain energy are significantly different in the two cases.

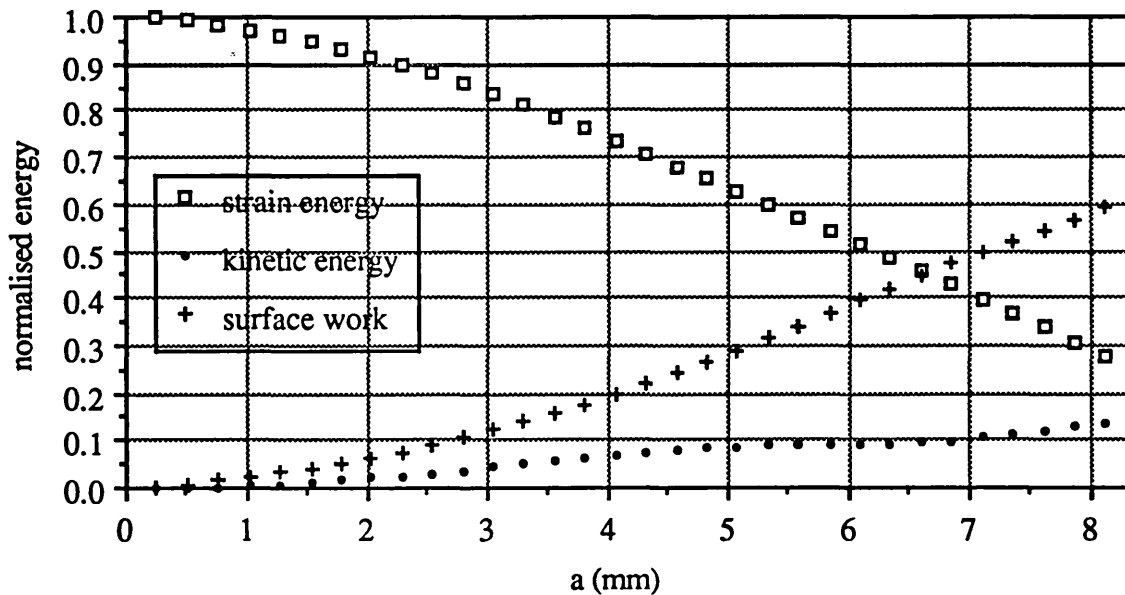


Figure 6.2.1.1: Normalized energies vs crack length increment, test PM4-2, H=40 mm, influence of the specimen height, (energies normalized by initially stored strain energy)

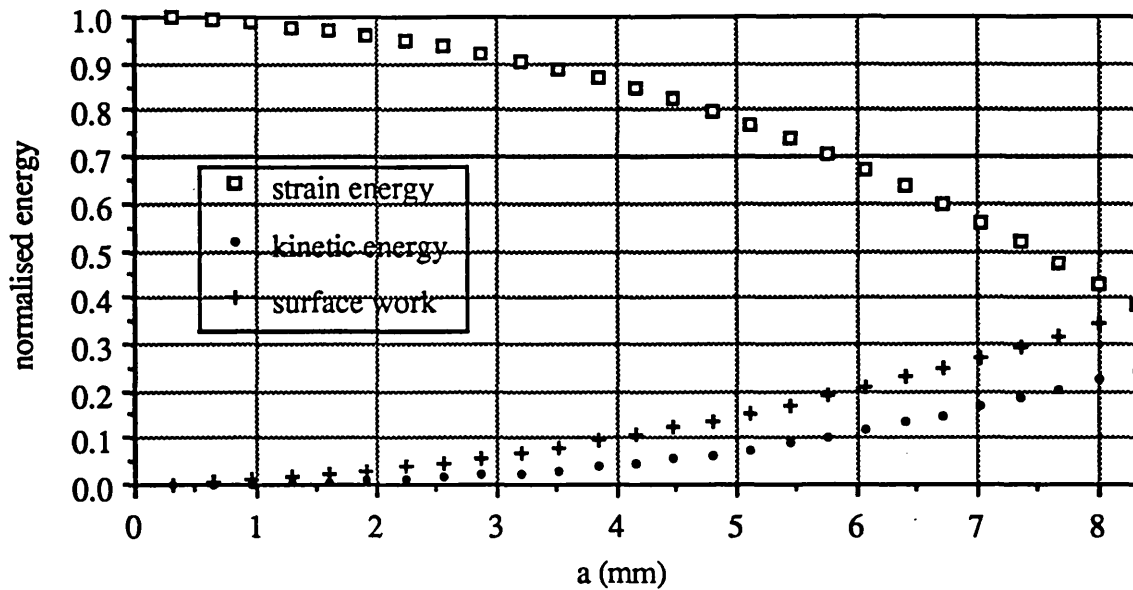


Figure 6.2.1.2: Normalized energies vs crack length increment, test PM7-1, H=80 mm, influence of the specimen height, (energies normalized by initially stored strain energy)

It might be expected that G^{dyn} should show some dependence on geometry and crack length, in addition to the crack speed, because it is the rate of release of stored energy from the sample of finite dimensions. If, however, the fracture resistance (G_{ID}) is to be a material property, it is difficult to see any physical explanation for the geometry dependence of G_{ID} . In support to this it should be pointed out that, in specimens with different heights, similar fracture surfaces were associated with RCPs at the similar speeds. If the energy is to be conserved, the equality between the crack driving force and the crack resistance (equation 1.2.2.6) should be maintained throughout RCP. Kalthoff[83] suggested that a minimum fracture energy requirement may exist, but if more is available than is needed, the material is able somehow to absorb more (section 1.4). This means practically that the minimum G_{ID} obtained from all

geometries at a given speed should be accepted as the only meaningful material parameter. Shockey and Kalthoff [83] speculated that at a high crack speed the equality between the crack driving force and the crack resistance does not hold i.e. that $G^{\text{dyn}} \geq G_{\text{ID}}$. This suggestion is in contradiction with the basic energy conservation principle and is difficult to accept (section 1.4).

Genussov[89] explored the possibility that a large scatter of several hundred percent in his G_{ID} results of PE was due to improper presentation of the mechanical properties of the polymer. He included a VE model in the analysis of RCP (section 5.2.3), and reported that the values of G_{ID} generated by the elastic model with single dynamic modulus were less than 10% higher than those generated by the VE model. His calculation showed that the amount of viscous energy, dissipated during RCP in the glass-PE SENT sample under constant-displacement conditions, was less than 4% of the energy absorbed by the crack. Without resolving the enigma of the highly scattered G_{ID} with crack length for samples fractured at constant speed, and with uniform fracture surface, Genussov[89] accepted the 'lower-shelf' values of G_{ID} (1-2 kJ/m²) as a material property - 'worst case' engineering practice. As a possible explanation for the crack reluctance to change speed in line with G^{dyn} at a given temperature, he supported the idea suggested by Ravi-Chandar and Knauss [84-b, 85] that the crack propagation speed depends only on the initiation event, and is largely independent of subsequent stress and the remote loading conditions (section 1.4). Ravi-Chandar and Knauss [84-b, 85] used an electrical method to rapidly load the crack face of a large SEN specimen made of Homalite-100. By changing the duration and the rate of the short pressure pulse, they obtained different crack speeds. While the crack propagated at constant speed, a large variation in K^{dyn} was observed along the specimen. The method of caustics was employed for direct K^{dyn} measurement. They explained this phenomenon in terms of a microcracking model, by speculating that the rate of microcracks coalescence established at initiation, remained constant during the fracture event. They concluded that: "The velocity with which the crack propagates is determined by the stress intensity factor at initiation". The crack propagation process was described as a process of linking up of many microcracks in front of the main crack. This involves a requirement to

accumulate a certain level of damage at the crack tip before initiation. Despite the fact that under gradually changing K^{dyn} the crack speed remained constant along the sample, a permanent step change in crack speed was produced by reflected stress waves (Ravi-Chandar and Knauss [84-b]). This was explained by speculating that the sudden change in K^{dyn} , due to a stress wave, altered the distribution of the microcracks at the crack tip.

The constant speed crack propagation results, reported in chapters 3 and 4 of this work, were obtained from tests on SENT specimens under fixed-grip conditions (as in Genussov's work). The specimens were loaded in tension by applying a uniform displacement along the boundaries, until RCP occurred. This uniformly applied displacement resulted in a rather uniform stress along the specimen ligament (figure 6.2.1.3), followed by the crack propagation at constant speed. Figure 6.2.1.3 shows the distribution of the nodal forces in y-direction (F_y) along the crack path prior to the crack initiation, as obtained from the FE simulation (test PM5-1, sections 4.4.1, 5.3.3). These forces are proportional to the normal stress σ_{yy} along the sample. Also shown in figure 6.2.1.3, are recorded crack speed data (section 4.4.1).

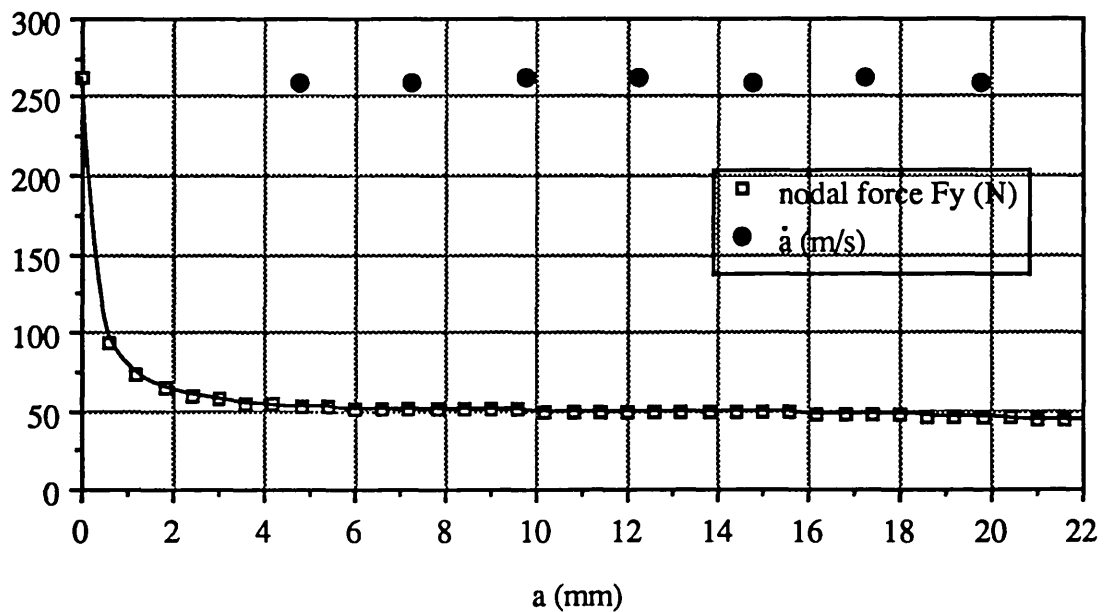


Figure 6.2.1.3: Fixed-grip uniform displacement test, influence on the crack speed

A simple test on the PMMA SENT configuration is performed to examine whether the changes in the crack speed can be provoked by altering the distribution of the stress (σ_{yy}) along the specimen ligament. Non-uniform stress distribution was produced using a particular method of loading: the specimen was partially gripped along the boundaries, and a uniform displacement was applied along the clamped regions (figure 6.2.1.4). The dimensions of the specimen were such that the eventual change in the crack speed could only be caused by the non-uniform stress distribution, as set by the remote loading conditions, and not by the sudden stress wave interaction with the crack tip. Two different specimen heights, 40 and 80 mm, were employed to obtain different levels of the stress non-uniformity along the sample.

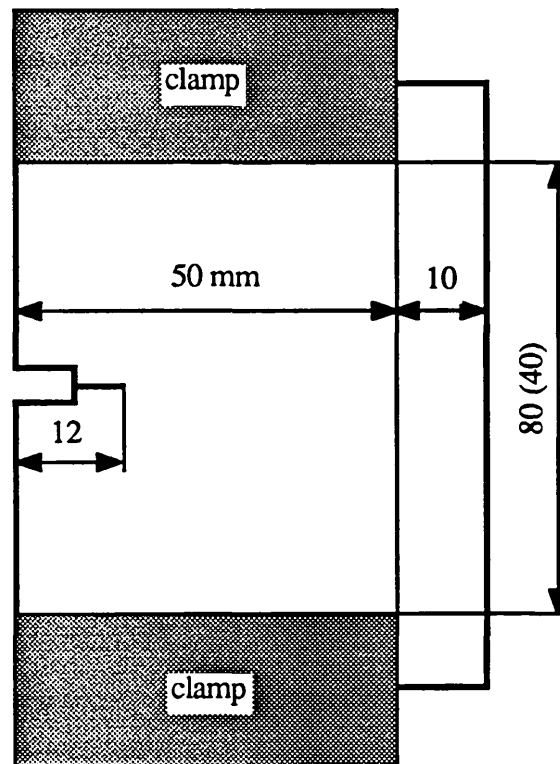


Figure 6.2.1.4: The configuration of the partially clamped PMMA SENT sample

The distribution of the nodal forces (F_y) along the crack path at the instant of the crack initiation, as set by the 'partially-clamped' boundary conditions, are presented in figures 6.2.1.5 and 6.2.1.6, for 80 and 40 mm tall specimens respectively.

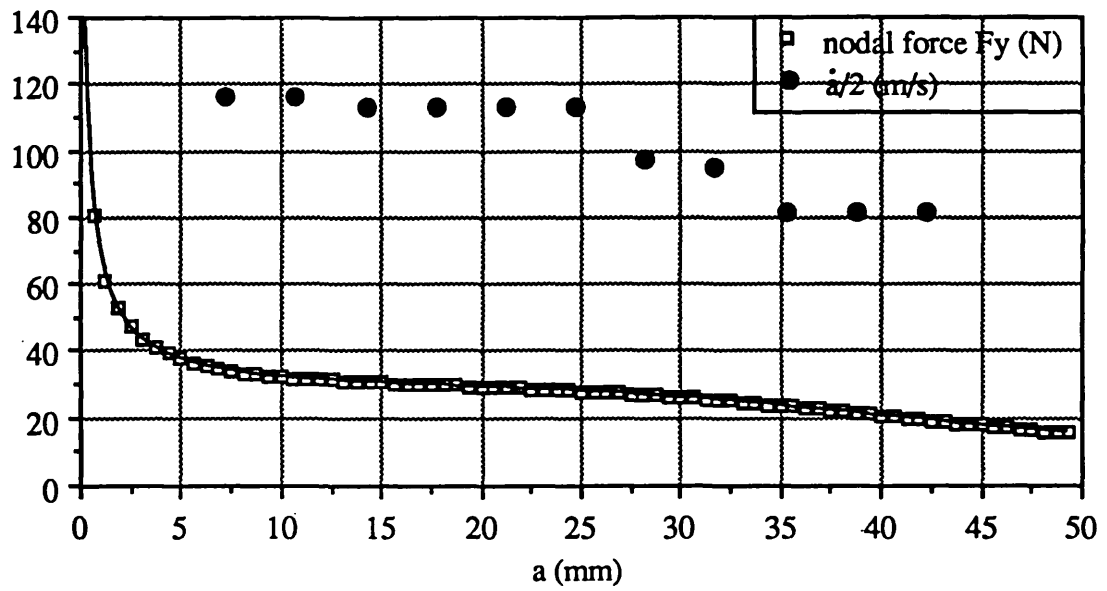


Figure 6.2.1.5: 'Partially-clamped' boundary conditions test, influence on the crack speed

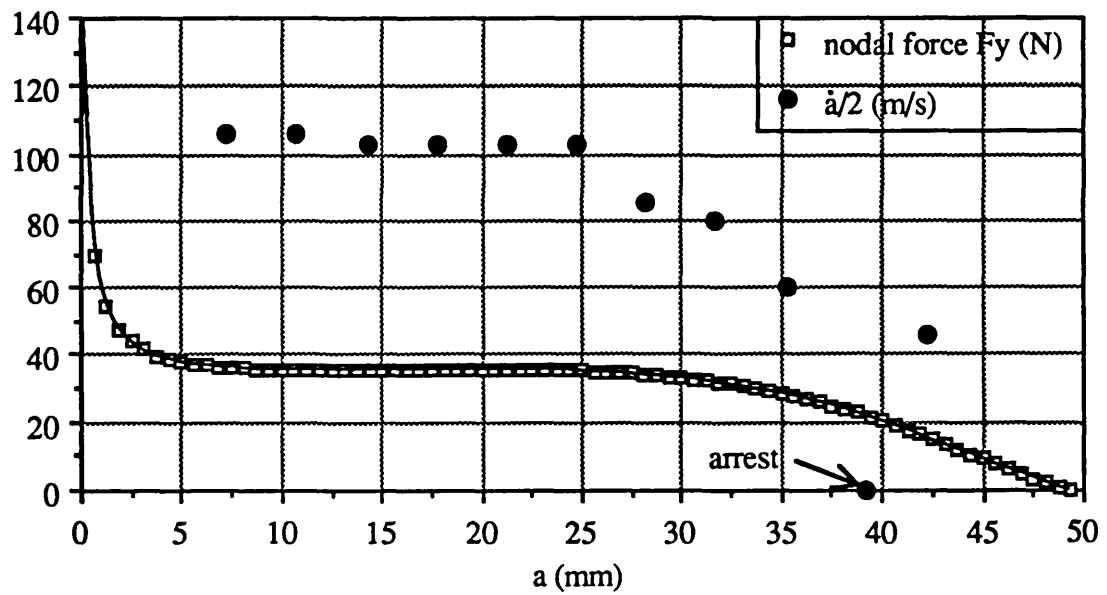


Figure 6.2.1.6: 'Partially-clamped' boundary conditions test, influence on the crack speed

This method of loading resulted in a reduction of the stress σ_{yy} over the last 23 mm of the specimen ligament. The changes in the crack speed along the specimen corresponded to the changes in σ_{yy} (figures 6.2.1.5-6). The details of the measurement of the crack length history were given in sections 4.3 and 4.4.1. A more severe reduction in σ_{yy} occurred in the shorter samples ($H=40$ mm). Consequently, larger changes in crack speed over the last 23 mm of the crack advance happened in these samples. Crack arrests and reinitiations were also observed in 40 mm tall specimens (figure 6.2.1.6). According to the crack speeds and velocities of the longitudinal waves, the first sudden interaction of the moving crack tip and the reflected stress waves occurred between 6.5 and 9.5 mm of crack advance in the 80 mm tall specimen, and between 3. and 4.2 mm in the 40 mm tall sample. This was followed by repeated wave arrivals at the crack tip. Thus, the changes in crack speed which took place beyond the crack length of about 27 mm, can be entirely attributed to the changes in the stress along the sample due to the loading conditions. Crouch[86] claimed that: "in the larger tension specimens and DCB geometries, inertia effects on the bulk motion of the specimen appear to predominate over superimposed stress waves influences, while in the low height strip geometry SEN tests, a steady-state of multiple boundary reflections is rapidly established".

The analysis of the constant speed RCP results, based on the single constant modulus, predicted large variations of G_{ID} (or K_{ID}) along the crack path (sections 5.3.2, 5.3.3, 5.3.4). It implies that during the crack propagation at constant speed in a SENT sample, large variations in normal stress σ_{yy} occurred along the ligament. This is in contradiction with the observations revealed above, which demonstrate that the crack speed is expected to change in line with the variation in the stress distribution along the crack path.

6.2.2 The sources of possible errors in the elastic analysis

The discussion from the previous section led the author to believe that G_{ID} vs a ($\dot{a}$) results given in chapter 5, do not properly characterize the constant speed RCPs in the duplex PMMA/PC, the PMMA and glass-PC SENT geometry. In this respect, it would be unwise and unjustifiable to accept that the computed G_{ID} results genuinely reflect the nature of the crack resistance of the analysed materials.

The accuracy of G_{ID} data, generated by the indirect method employed in this work, relied on the accuracy of both the experimental techniques and the numerical simulation. Thus, the errors which may be responsible for the improper characterization of G_{ID} values, could have been generated by:

- a) the errors involved in the measurement of the breaking load (P_b),
- b) the inaccuracies in the recorded crack tip position during RCP,
- c) the 'unrealistic' simulation of the conditions of the loaded boundaries during RCP,
- d) the improper representation of the mechanical properties of the materials.

The error involved in the monitoring of P_b (section 3.4), is estimated to be less than $\pm 1\%$. The details of the measurement of the crack length histories for various geometries were presented in chapters 3 and 4. The sensitivity of the FE simulation of the RCP to the inaccuracies involved in the specified crack length vs time data, was examined in chapter 5. It can be concluded that the unacceptably large variations of G_{ID} vs a ($\dot{a}$) were not just a consequence of such inaccuracies.

Loaded boundary influence on G_{ID} results

The third factor (c) considers the accuracy of the simulation of the loaded boundary position during the crack propagation. Prior to the assessment of the error caused by the inaccuracies in this simulation, a simple numerical test is presented below, aiming to illustrate the influence of the conditions of the loaded boundary during RCP on G^{dyn} results. A SENT sample with the assumed dimensions ($D \cdot H \cdot B = 12.56 \cdot 100 \cdot 10$ mm), material properties ($E = 3$. GPa, $\nu = 0.33$, $\rho = 1180$ kg/m³), breaking load ($P_b = 1.5$ KN) and the crack speed ($\dot{a} = 343$ m/s), is analysed numerically by modelling two extreme cases of loaded boundary conditions during fracture:

- 1) fixed-displacement conditions of the upper boundary,
- 2) free (non-constrained) conditions of the upper boundary.

The initial static solution, corresponding to the boundary load (P_b) prior to the crack initiation, is identical for both simulations. In the second simulation, the slow stretching is followed by the instantaneous relaxation of the sample at the instant of the crack initiation, i.e. all upper boundary nodes were suddenly set free at that instant. Both simulations resulted in identical G^{dyn} data for the first 7 mm of crack advance (figure 6.2.2.1). As the compressive stress wave, generated by the sudden relaxation of the loaded boundary (simulation -2), reaches the moving crack tip the crack becomes aware of the sample height and the conditions of the top surface. After this moment, G^{dyn} values from the second simulation starts to diverge from the results of the first simulation, where RCP is completed before the stress wave, generated at the crack tip at initiation (section 6.2.1), returns from the loaded boundary to the moving tip.

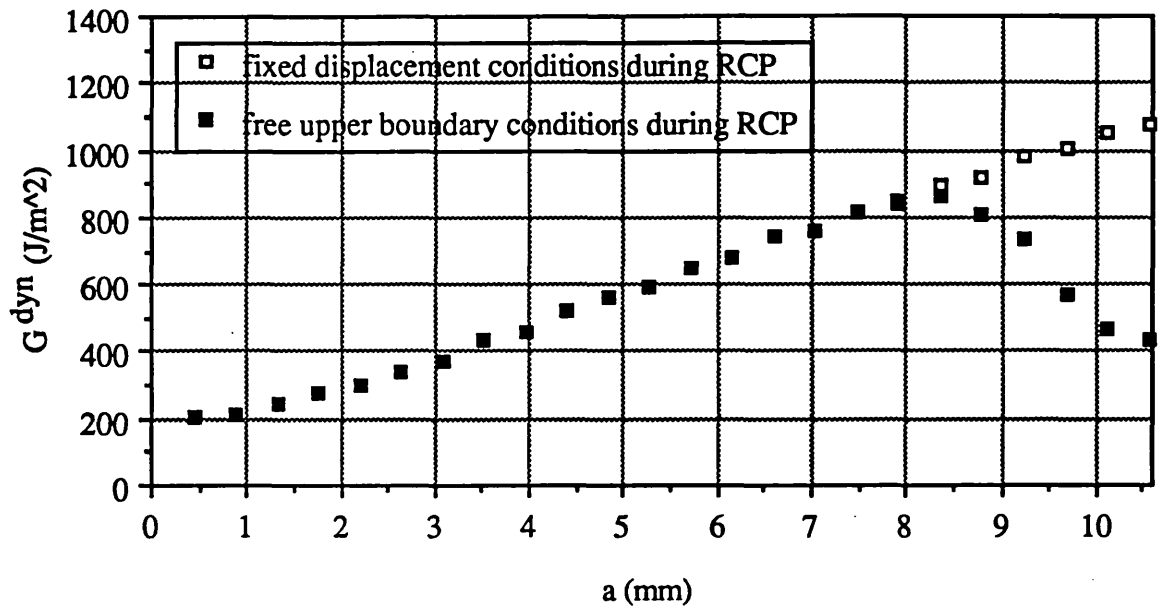


Figure 6.2.2.1: Crack driving force vs crack length increment,
influence of the conditions of the loaded boundary during RCP, on G^{dyn} results

Each tested SENT specimen (chapters 3, 4) was clamped 'rigidly' between two heavy grips and loaded slowly in tension until the crack initiated. During the fast crack propagation, the setup was regarded as being under fixed-displacement conditions (sections 3.2, 5.3.1). However, if the grips move during RCP, a certain amount of the energy stored in the specimen will be transferred to the grips as kinetic energy. Consequently, the crack would have less energy available to be driven forward. An assessment of whether the loaded boundaries are likely to remain fixed during RCP is considered below.

Numerical simulation, which allows for the motion of the clamped boundary during RCP, is necessary in order to examine the influence of the conditions of loaded boundary on G_{ID} results. First of all, the machine stiffness and the weight of the loading grips have to be

determined. The machine stiffness was determined by loading an uncracked steel sample, which was clamped between the same grips that were used throughout this work (chapters 3 and 4). The sample was loaded in tension up to 10 KN. The loading-unloading cycle was repeated several times. The loading rate of 10 mm/min was retained during the test. The displacement of the steel sample was measured by a Linear Voltage Displacement transducer. Load-displacement curves were recorded on a Hewlett Packard X-Y plotter. The value of the machine stiffness, averaged from about 20 measurements, was found to be 1×10^8 N/m. The weight of the grips used in this work was 14.35 kg each.

The numerical simulation of the 'realistic' conditions of loaded boundary during RCP is done as follows:

a) The mass of the grip is equally distributed on the upper boundary nodes of the mesh (due to symmetry, only half of the sample is modelled). The machine loading system was simulated by linear springs, attached to each upper boundary node in the vertical direction.

b) The deformation of the loading system in the vertical direction is assumed to be uniform during the test, i.e. the loaded boundary of the sample remains horizontal at any instant of time. Thus, as before, the initial static loading is simulated by applying a uniform displacement (VD) to the upper boundary nodes in the vertical direction, until the total applied load is equal to the measured load (P_b) prior to fracture (section 5.3.1). This implies that each spring is equally stretched in the vertical direction. This initial deformation of the springs (VSI) is determined by equation 6.2.2.1, according to P_b and the measured machine stiffness K_M :

$$VSI = \frac{P_b}{K_M} \quad 6.2.2.1$$

where:

$$P_b = \sum_{i=1}^N FTY_i$$

FIY_i are the initial loads acting on the upper boundary nodes in the -y- direction prior to fracture, and N is the number of nodal points on the upper boundary (and hence the number of springs).

c) The stiffnesses of the springs applied on the upper boundary nodes are evaluated from equation 6.2.2.2:

$$FIY_i = K_i VSI \quad (i = 1, N) \quad 6.2.2.2$$

where: K_i is the stiffness of the i^{th} spring. Note that K_i is not constant, and that:

$$\sum_{i=1}^N K_i = K_M$$

d) The program switches to the dynamic mode at the instant of the crack initiation, i.e. at the instant the initial crack tip node is suddenly released (section 5.3.1). During the dynamic mode simulation, the 'force boundary condition' is applied to the upper boundary nodes, according to the elastic deformation of the springs (equation 6.2.2.3). At the instant of crack initiation, these forces (FY_i) are equal to the initial loads (FIY_i).

$$FY_i = K_i (VSI - (V - VD)) \quad (i = 1, N) \quad 6.2.2.3$$

where: V is the momentary displacement of the upper boundary in the -y- direction.

This enabled the modelling of the 'realistic' conditions of the top surface of the specimen during RCP. However, the analysis of the experimental results showed that the variation of the upper boundary position with time during fracture, and thus the energy transferred into the clamps, were practically negligible. Due to the high inertia of the clamp and the high machine stiffness, no motion of the clamp was detected and the loaded boundary remained fixed during

the crack propagation. It was already pointed out (section 6.2.1), that the calculated G_{ID} cannot be affected by the way the clamped boundary was simulated until the first stress wave returns from that boundary to the crack tip (regardless of the grip mass and the stiffness of the machine). Obviously, the anomalies in calculated G_{ID} vs a (Å) data are not a consequence of the 'unrealistic' simulation of the conditions of the loaded boundaries during RCP. Therefore, this problem cannot be resolved simply adjusting the boundary conditions. Similar observations were reported by Crouch[86]. Analysing a strip geometry tension tests he allowed for the rotation of the loaded boundary during initial loading, and for the motion of that boundary during RCP. Both the rotation and the motion of the loaded boundary were found negligible.

A simple experiment on the PMMA SENT configuration ($D/H = 0.375$) was performed in order to check the accuracy of the numerical prediction. An accelerometer, type 4369, product of Bruel & Kjaer - Denmark, was used to record the movement of the clamp during RCP. It was placed on the top of the upper clamp. The grid line technique was employed for measurement of the crack length history. Both the accelerometer signal and the timing line signal were recorded on a digital oscilloscope. The recording was triggered by the breaking of the conductive line, painted in front of the crack tip (section 4.3). As predicted by the FE simulation, no motion of the clamp was observed during the crack propagation. Using the strain gauges placed near the loaded boundary of the steel SENT specimens, Dahlberg et al [80] proved that the fixed boundary conditions were fulfilled during RCP (section 3.2).

The single constant modulus

Based on the discussion given in section 5.2.4, the behaviour of the tested polymers during RCP was represented with the single constant dynamic modulus (E_d). However, the effect of different moduli on G_{ID} results is considered here. An experiment on the PMMA SENT sample is simulated by employing firstly the upper bound modulus - E_d , and secondly the

lower bound modulus - E_s (table 2.2.1). The details of the sample geometry, breaking load and the crack speed are given in table 6.3.2.1 (sample SG1-1/4). In both cases the initial slow loading is modelled by uniformly deforming the loaded boundary until the total applied load reaches the measured breaking load - P_b . Thus, the amount of the strain energy available for crack propagation is substantially larger in the static than in the dynamic modulus analysis. Consequently, the crack resistance values obtained when using the static modulus are higher than the dynamic modulus results (figure 6.2.2.2).

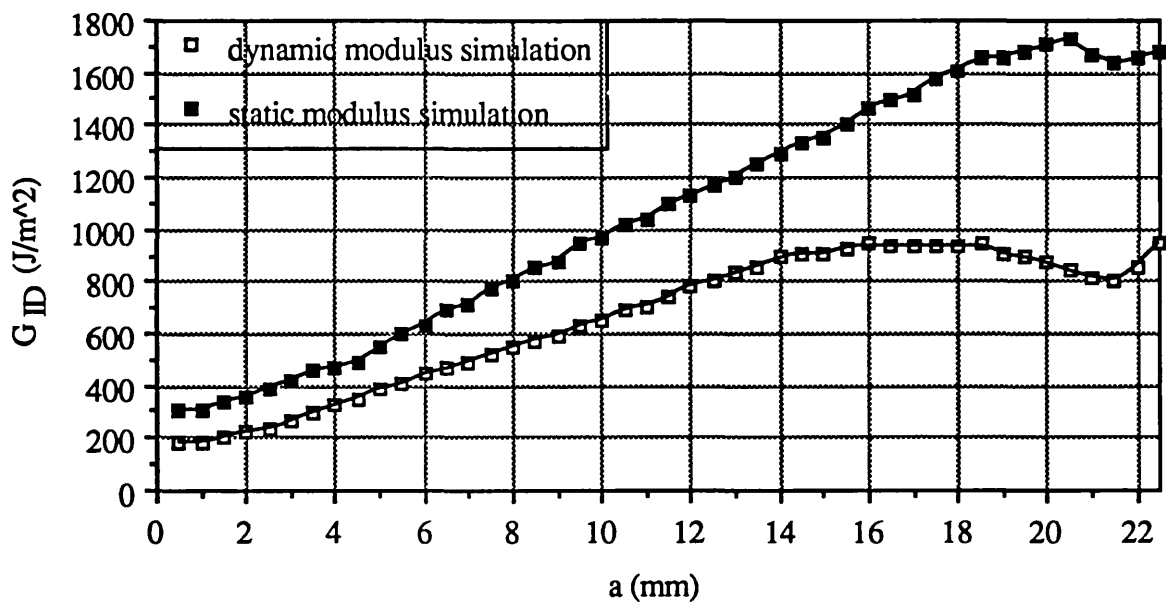


Figure 6.2.2.2: Crack resistance vs crack length increment; influence of various single moduli

Although the crack propagated at constant speed leaving a uniform fracture surface behind, both cases predicted large variations of G_{ID} values with crack length (figure 6.2.2.2). The static and the dynamic modulus analyses generated the upper and the lower bound G_{ID} results, respectively. Any other single constant modulus simulation will result in G_{ID} data lying between these bounds, and hence in a large variation of G_{ID} with crack length and speed.

6.3 Local strip model

6.3.1 Introduction

Having examined the effects of various factors involved in the characterization of the crack resistance (sections 6.2.1, 6.2.2), the question arises of whether the linear elastic analysis, with a single constant modulus, can be used to model RCP adequately. The FE simulations of constant speed RCPs, employing the single constant modulus (chapter 5), predicted large variations of the normal stress σ_{yy} (and thus the strain ϵ_{yy}) in the vicinity of the moving crack tip along the sample ligament. Consequently, K_{ID} and G_{ID} varied with the crack length (equations 1.2.2.1, 1.2.2.5). Quite contrary, section 6.2.1 demonstrated that the crack speed should change in line with the stress variation along the crack path. Further examination of the FE prediction of the strain and stress variations with crack length during RCP is therefore necessary. According to fracture surfaces of specimens broken at constant crack speeds (sections 3.5, 4.4.1 and 4.4.2), the crack resistance, and thus σ_{yy} and ϵ_{yy} near the moving crack tip, are expected to be rather independent of the crack length (section 6.2.1).

6.3.2 Measurement of the strain along the crack path during RCP

Measurement of strain histories - ϵ_{yy} vs t (ϵ_{yy} is a normal strain in respect to the crack plane), along the crack path during RCP, was performed on a number of the PMMA and glass-PC SENT specimens. Two strain gauges (gauge type EA-06-060LZ-120, gauge length 1.5 mm, product of Micro Measurements - USA) were adhered on each sample near the prospective crack path (figure 6.3.2.1). The choice of the strain gauges, the preparation of the specimens and the bonding of the strain gauges were made according to the instruction manuals of Micro Measurements Division: Catalog 500 and Bulletin 309B. During the crack propagation, the normal strain (ϵ_{yy}) was recorded as a function of time at two positions (figure 6.3.2.1). Similar measurement of strain histories during RCP were reported by Dally and Sanford [89], Sanford et al [90], Sun[83], (section 2.3.1).

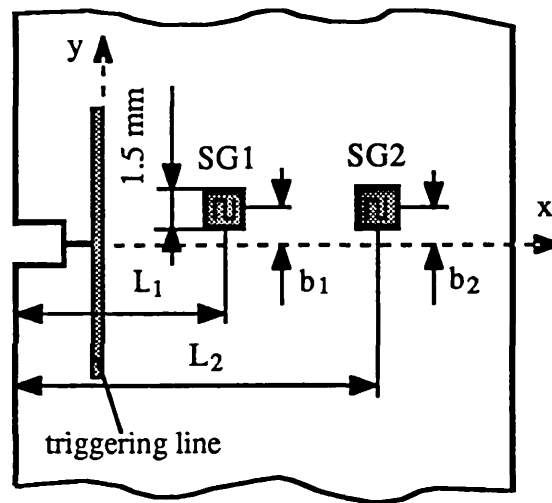


Figure 6.3.2.1: Strain measurement configuration

The position of the strain gauge - SG1, which was placed closer to the initial crack tip, is defined by L_1, b_1 , while the position of SG2 is defined by L_2, b_2 (figure 6.3.2.1). As the moving crack tip approaches the strain gauge, measured strain is expected to reach its

maximum value - peak. The average crack speed can be evaluated from the distance between the strain gages (L_2-L_1), and the time difference between the peaks. The signals from both strain gauges were monitored by a fast digital oscilloscope - Gould 1604, 20 MHz. The breaking of the conductive triggering line was used in order to start recording (section 4.3). A conductive, silver-paint line was painted right across the tip of the prenotch of the PMMA SENT sample (figure 6.3.2.1). In the case of the glass-PC specimen, a triggering line was applied on the glass slide (section 4.3). Thus, the breaking of the triggering line was caused either by the crack initiation in the PMMA sample, or by the fracture of the glass slide of the glass-PC sample. The oscilloscope was set on its fastest sampling rate, i.e. a resolution of 0.1 μ s per point was used. Each specimen was broken in tension under the same loading conditions as used throughout this study (sections 3.4, 4.3). Typical ϵ_{yy} vs t results of the PMMA and glass-PC SENT specimens are presented in figures 6.3.2.2 and 6.3.2.3 respectively. The details of the corresponding specimen dimensions, positions of the strain gages, crack speed and breaking load are listed in table 6.3.2.1. The crack speed was assumed to be constant throughout. According to uniform fracture surfaces of the broken samples, this should be a valid assumption (see chapter 4 and section 6.2.1).

material	sample	B	D	H	a_0+a_1	L_1	L_2	b_1	b_2	$\dot{a}$ (m/s)	P_b (KN)
PMMA	SG1-1/4	9.93	25.75	104.	3.4	10.	20.	1.3	1.25	348	2.83
PC	PCSG-2	9.35	30.3	60.	5.6	10.6	21.6	1.5	1.5	535	5.35

Table 6.3.2.1: Strain measurement-samples dimensions and SGs positions (dimensions in mm)

Zero time corresponds to the breaking of the triggering line for both tests (figures 6.3.2.2-3). While this moment coincides well with the crack initiation in the PMMA sample, the crack initiated in the glass-PC specimen about 10 μ s after the triggering occurred (figure 6.3.2.3).

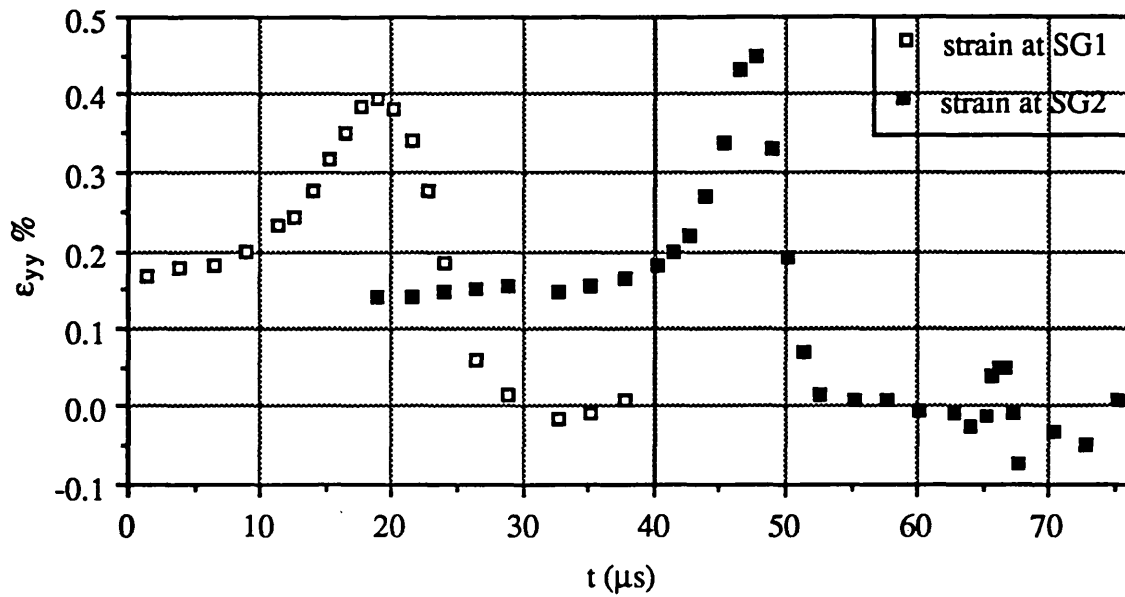


Figure 6.3.2.2: Strain vs time records from two strain gauges, test SG1-1/4

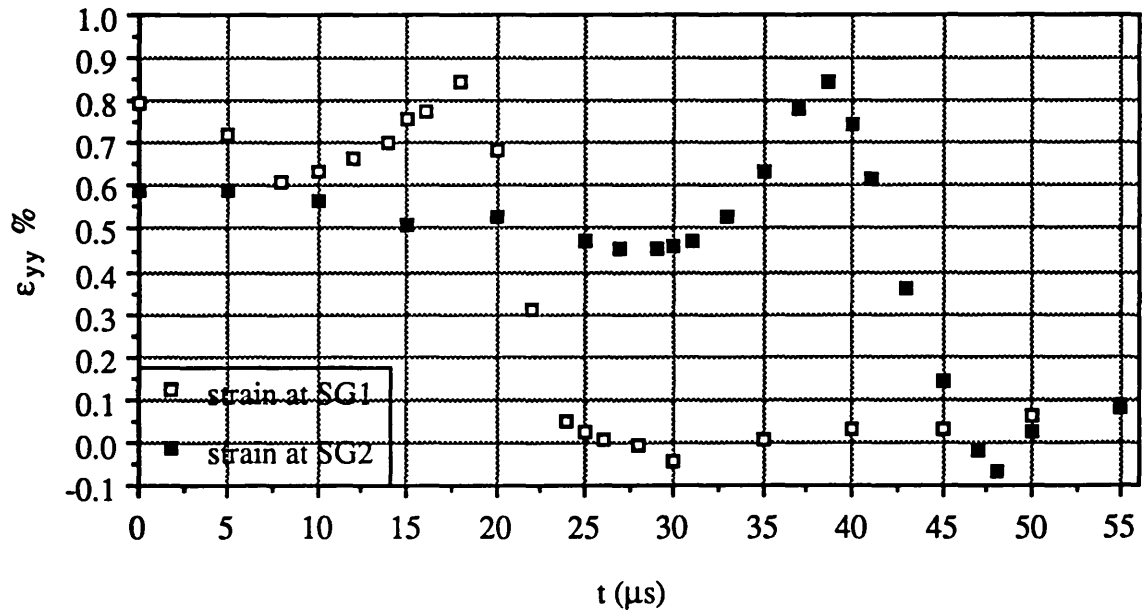


Figure 6.3.2.3: Strain vs time records from two strain gauges, test PCSG-2

The most significant observation regarding the experimental ϵ_{yy} vs t results of the PMMA and glass-PC SENT specimens is the very small variation of the peak strain over the distance L_1 to L_2 . For example, $\epsilon_{yy2}/\epsilon_{yy1} = 1.125$ for the PMMA sample, while $\epsilon_{yy2}/\epsilon_{yy1} = 1$ for the glass-PC sample (table 6.3.2.1, figures 6.3.2.2-3). Subscripts 1 and 2 refer to the strain gauges SG1 and SG2, or to the crack tip positions L_1 and L_2 , respectively. In the PMMA sample, the second peak strain was somewhat higher than the first one, probably because SG2 was closer to the crack path than SG1 (see table 6.3.2.1).

6.3.3 Comparison between the experimental and numerical ϵ_{yy} vs t results

The tests on the PMMA and glass-PC SENT specimens of previous section are numerically analysed in this section. Each experiment is simulated by using both single constant dynamic modulus - E_d , and single constant static modulus - E_s (table 2.2.1). According to the exact positions of the strain gauges SG1 and SG2 (figure 6.3.2.1), numerical ϵ_{yy} results are obtained by averaging the strain over a height of 1.5 mm - the height of strain gauges. At least three rows of elements were used to represent the height of the strain gauges. The comparison between the experimentally recorded and the numerically generated strain histories at the two positions along the crack path during RCP, is illustrated in figures 6.3.3.1 and 6.3.3.2. These show ϵ_{yy} vs t results of the PMMA sample SG1-1/4, and the glass-PC sample PCSG-2 respectively (table 6.3.2.1).

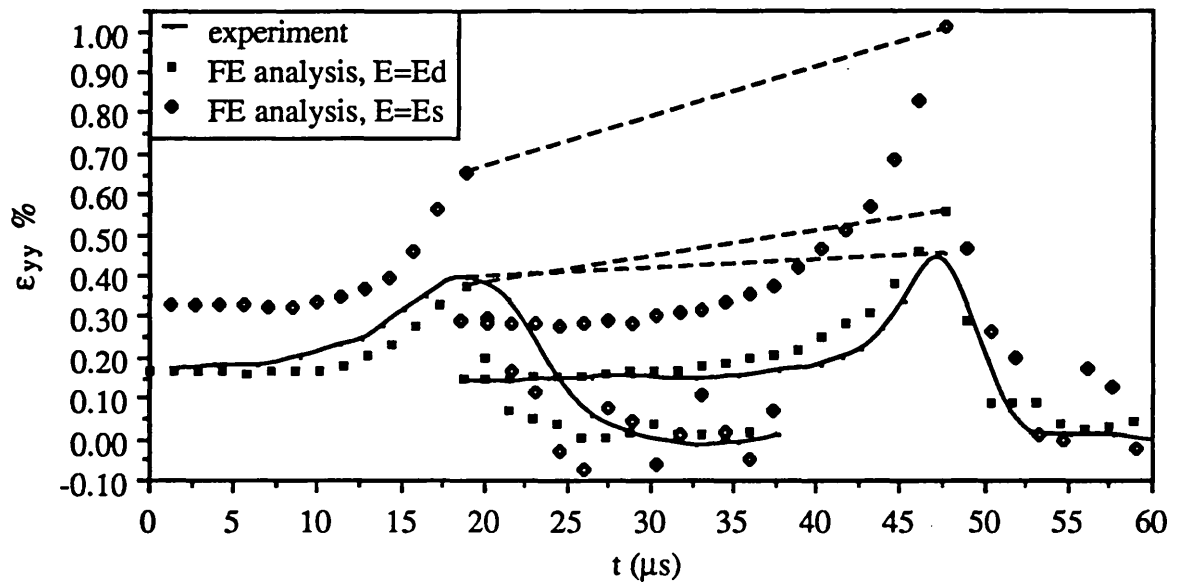


Figure 6.3.3.1: Comparison between experimental and numerical strain results, test SG1-1/4

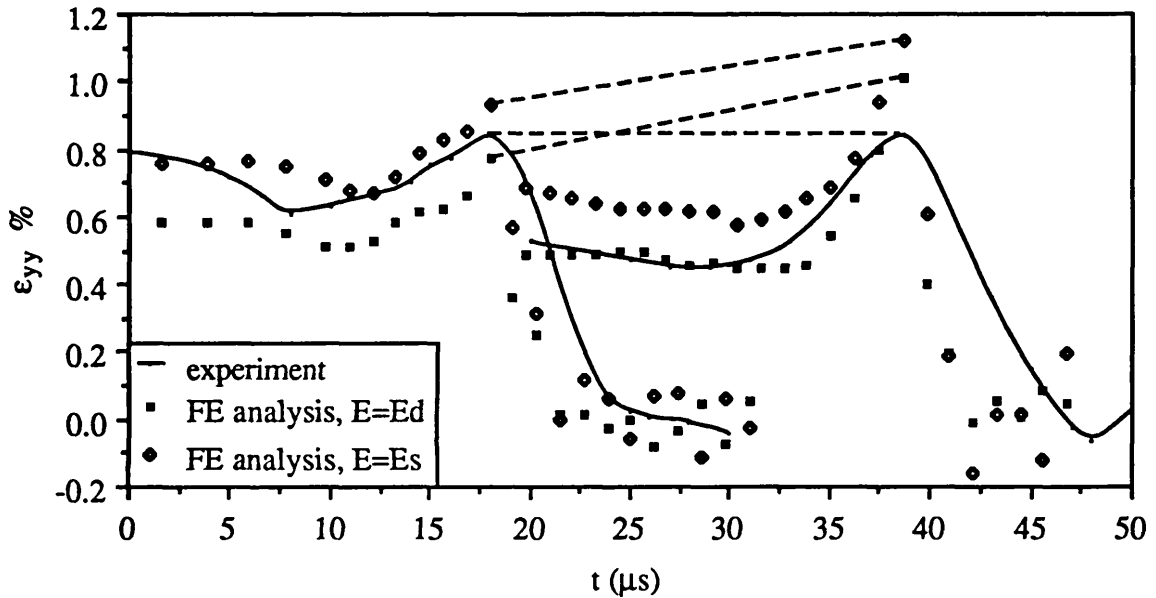


Figure 6.3.3.2: Comparison between experimental and numerical strain results, test PCSG-2

Broken lines in figures 6.3.3.1-2 approximate experimental and numerical peak strain histories.

The comparison between ϵ_{yy} vs t results obtained experimentally and by the FE analyses employing single constant modulus (figures 6.3.3.1-2), reveals several important conclusions:

a) All samples:

- When either single constant E_s or single constant E_d was used, the numerically predicted variation of the peak strains during the crack propagation from L_1 to L_2 , is considerably larger than the measured change of the peak strains over the same distance. $\epsilon_{yy2}/\epsilon_{yy1} = 1.489$ (1.545) from the single constant E_d (E_s) analysis of the PMMA sample, and $\epsilon_{yy2}/\epsilon_{yy1} = 1.3$ (1.2) from the E_d (E_s) analysis of the glass-PC sample (table 6.3.2.1, figures 6.3.3.1-2). ϵ_{yy1} and ϵ_{yy2} are the maximum average strains calculated during RCP at SG1 and SG2 respectively. These maximum strains correspond to the instants when the crack tip reaches the strain gauges. Both SG1 and SG2 were placed very close to the crack path (figure 6.3.2.1, table 6.3.2.1). Thus, ϵ_{yy1} and ϵ_{yy2} are the strains in the local region near the tip of the moving crack at lengths L_1 and L_2 respectively. As the crack advances from L_1 to L_2 the calculated strains (and stresses) in the local crack tip region increase (figures 6.3.3.1-2). Hence, the differences between the calculated and measured local strains develop over the given distance, and so the errors in predicted local strains. Therefore, the errors in calculated K^{dyn} develop accordingly (K^{dyn} is a measure of the local stress field, equation 1.2.2.1), while the errors in the generated G^{dyn} 'square-up' over the distance L_1 to L_2 (equation 1.2.2.5). Obviously, the analysis using the single constant modulus, either E_d or E_s , cannot adequately predict the variation of the local strain near the crack tip, during RCP in the PMMA and glass-PC SENT samples. Thus, generated G_{II} values cannot be accepted as genuine. The above observation shows that the assumed linear elastic material behaviour around the rapidly moving crack tip is not justified. This is a crucial point in the analysis.

b) PMMA SENT samples:

- ϵ_{yy} vs t results from the numerical analysis based on the single constant E_s appear to be higher than the recorded strains (figure 6.3.3.1). Neither the strains measured when the approaching crack tip was far from the strain gauge, nor the measured peak strain when the crack reached the gauge are reproduced.

- ϵ_{yy} vs t data obtained from the numerical simulation with the single constant E_d compare reasonably well with the strains measured during the crack propagation towards the strain gauges (figure 6.3.3.1). Thus, except for the local region in the vicinity of the moving crack tip, the choice of the short term modulus (E_d) for the analysis of RCP in the PMMA SENT specimens is satisfactory.

c) glass-PC SENT samples:

- Strain values, measured as the crack tip was approaching the strain gauges, lie between ϵ_{yy} vs t values generated numerically with the single constant E_d and E_s (figure 6.3.3.2). Thus, except for the local zone near the moving crack tip, use of either E_d or E_s should be adequate for modelling of RCP in glass-PC SENT samples. G_{ID} vs a data computed employing either single constant E_d or E_s are in a very close agreement (figure 6.3.4.5).

6.3.4 Local strip with low modulus

It was demonstrated above that the critical source of error in the analysis is the improper representation of the material behaviour around the rapidly propagating crack tip. It has already been recognized that any fracture event is dominated by the stress-strain-strain rate relation of the material. This relation governs both the local process at the crack tip and the capability of the remote stress field to transmit its energy into the region where the crack propagation takes place (Genussov[89]). Kanninen[85] acknowledged that non-uniqueness in G_{ID} vs $\dot{a}$ relationships could result from the local non-linear effects around the crack tip (section 1.4). Ravi-Chandar and Knauss [84-b, 85] argued that microcracks, distributed throughout the process zone of the moving crack tip, set a low effective stress wave speed (section 1.4). This suggests a low modulus in the local crack tip region. Freund and Lee[90] studied the process of dynamic tensile crack growth under small scale yielding conditions. The crack tip plastic zone was modeled as a strip zone extending ahead of the advancing crack tip. They observed that the dependence of the crack driving force on the crack speed could be markedly affected by the material inertia in the crack tip plastic zone region. Achenbach et al [79] noticed that the reduction in the slope of the stress-strain curve significantly increases dynamic effects in rapid crack propagation. Although the crack speed could be small when compared to the wave speed in a region away from the crack tip, it can be a considerable fraction of the wave speed at high strains near the tip. Broberg [71, 73] pointed out that "a neighbourhood of a crack tip, running or not, is always a region of instability", and that "In the dynamic case this fact implies that the propagation velocities of irrotational and equivoluminal waves tend towards zero, irrespective of plastic flow being present or not". He ([73]) studied the influence of the low wave speed near the crack tip by using a thin strip of low elastic modulus along the crack path surrounded by a higher modulus infinite medium. The modulus of each region was constant and the crack speed was assumed higher than the longitudinal wave speed in the layer, but smaller than the equivoluminal wave speed in the surrounding medium outside the layer. It was found that the

singularity at the tip of the crack vanished, and the crack tip blunted. Removal of the singularity was noted to imply that the crack extension might not occur exclusively from the tip and, instead, irregular crack propagation and rough fracture surfaces might be induced. The influence of the low wave speed in the thin layer around the crack tip was found substantial. The energy flow to the crack tip was greatly reduced when compared to the homogeneous case. Broberg argued that neglecting the energy dissipation in the eventual plastic region was not critical, because, the mechanical processes during loading are equal in both elastic-plastic and non-linear elastic medium, and unloading takes place behind the crack tip. Williams and Ivankovic [91] used the Spring-Mass model (see appendix A2) to determine a crack speed, crack length and loading relationship in terms of energy release rate. The analysis was used to describe crack speed data taken over a wide range of geometry changes and a good fit was obtained by predicting a much lower wave speed than expected. This was rationalized on the basis of a low local modulus caused by large strains in the vicinity of the crack tip.

Accurate simulation of RCP requires an appropriate stress-strain-strain rate (-temperature) relationship of the material under consideration. Unfortunately, the parameters of the materials involved in this work which may describe the material behaviour near the rapidly moving crack tip, are uncertain because of large strains and high strain rates (and high temperature). Due to the project's 'time constraint', the determination of relevant constitutive relationships was not possible. Also, the modelling of the eventual non-linear material behaviour increases the complexity of the analysis as well as the computational cost. Section 6.3.3 demonstrated that the main source of error, in the characterization of RCP parameters, was the improper representation of the local material behaviour. It further showed that the use of E_d in the PMMA and either E_d or E_s in glass-PC SENT samples were satisfactory for modelling of the material behaviour away from the crack tip during RCP. Considering the above arguments, the material behaviour around the crack tip was approximated using a local strip along the prospective crack

path (figure 6.3.4.1). The crucial characteristics of this simple model should be much lower stress wave velocities around the crack tip than at larger distances. Broberg[73] claimed that "The presence of the layer should not be very important except for the neighbourhood of the crack tip". He concluded that the essential features of RCP in a non-linear medium were satisfactorily modelled by the local strip model. Unlike the local strip zone, a more sophisticated model should be able to simulate a more realistic local 'non-linear' zone which is advancing with the moving crack tip. However, this inability of the adopted model is not essential in the analysis (Broberg[73]).

In the present study, both the thin strip layer and the surrounding material are taken as isotropic and linearly elastic (as in Broberg[73]). The crack is propagated in the plane of symmetry of the strip (figure 6.3.4.1). The energy dissipation due to eventual plastic deformation is not considered by this model. As in the analyses with single constant modulus (chapter 5), calculations of RCP properties employing a local strip zone around the crack path are based on the magnitude of the breaking load, dynamic modulus outside the layer, recorded crack length histories and fixed-displacement conditions of the loaded boundary during RCP. Again, the equality between the crack driving force (G^{dyn}) and the crack resistance (G_{ID}) is assumed to be satisfied as long as the crack propagates (equation 1.2.2.6).

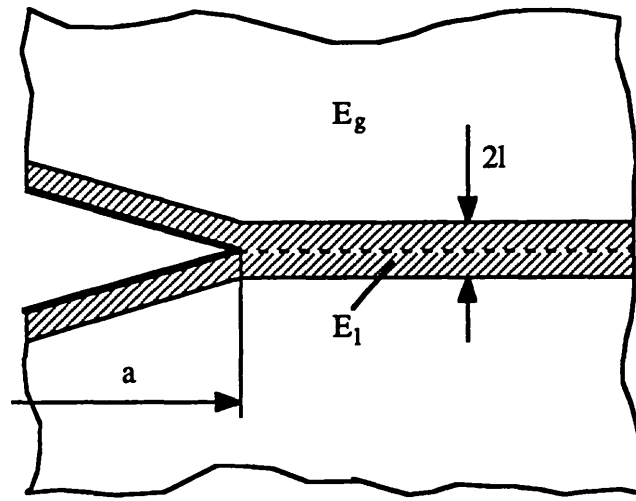


Figure 6.3.4.1: Local Strip Model Configuration, E_g - global modulus for the bulk material outside the layer, E_l - local modulus for the material inside the layer, $2l$ - the thickness of the local strip around the crack path, a is the crack length at given time

The question now arises of how to define the local strip parameters - E_l , l , (figure 6.3.4.1)? As revealed above, the local strip zone should adequately model the behaviour of the material in the neighbourhood of the crack tip during the crack propagation. Thus, the measured peak strains (their magnitude and the variation over the given distance) should be reproduced when an appropriate choice of E_l and l is made.

The tests on the PMMA and glass-PC SENT samples of section 6.3.2, are re-analysed using the local strip model. The comparisons between measured peak strains, the peak strains predicted by single constant modulus analyses and the best-fit predictions of the peak strains from the local strip analysis are given in figure 6.3.4.2 for the PMMA sample SG1-1/4, and in figure 6.3.4.4 for the glass-PC sample PCSG-2 (table 6.3.2.1, sections 6.3.2-3). Straight lines were drawn through the numerical peak strains ϵ_{yy1} and ϵ_{yy2} , for clarity purposes (figures 6.3.4.2 and 6.3.4.4). Corresponding G_{ID} vs a results are presented in figures 6.3.4.3 and 6.3.4.5, respectively.

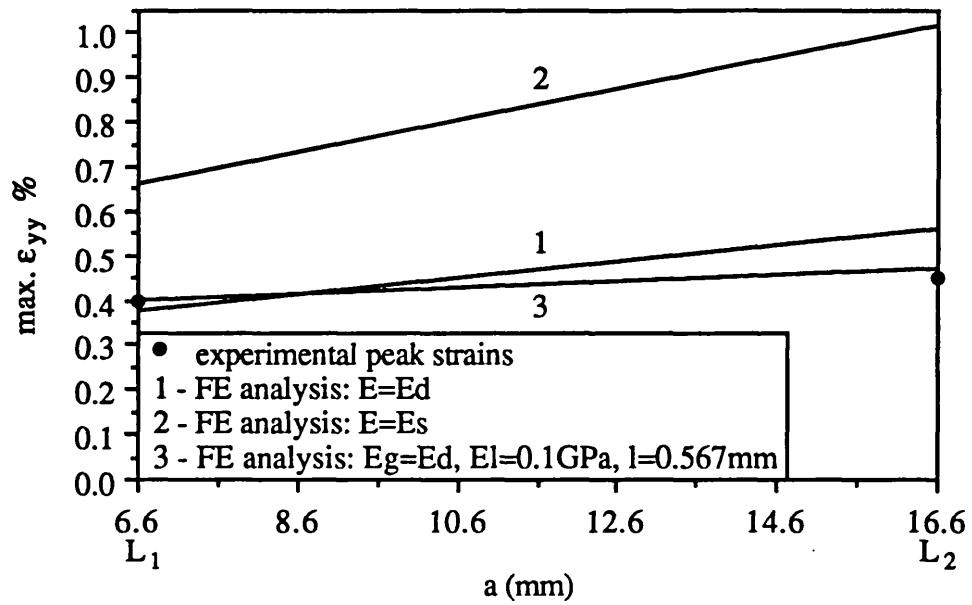


Figure 6.3.4.2: Peak strain vs crack length increment, test SG1-1/4

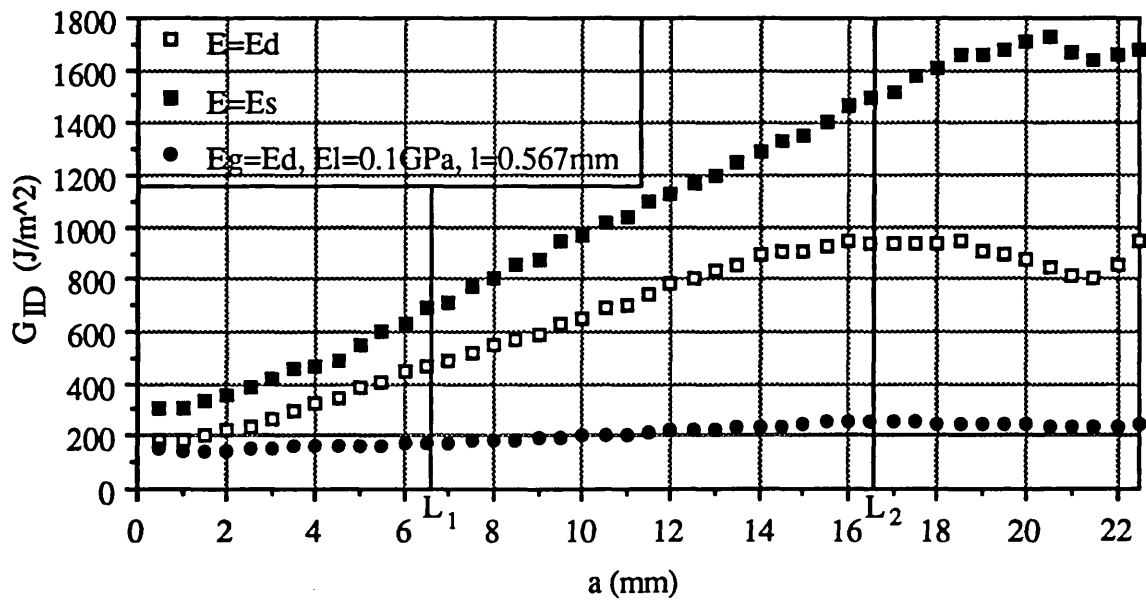


Figure 6.3.4.3: Comparison between G_{ID} results from different FE analyses, test SG1-1/4

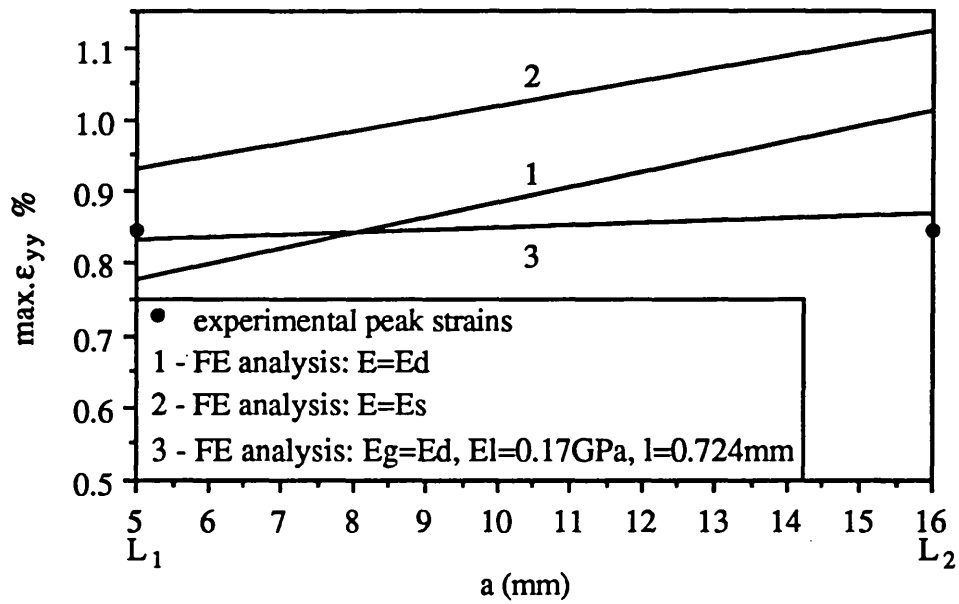


Figure 6.3.4.4: Peak strain vs crack length increment, test PCSG-2

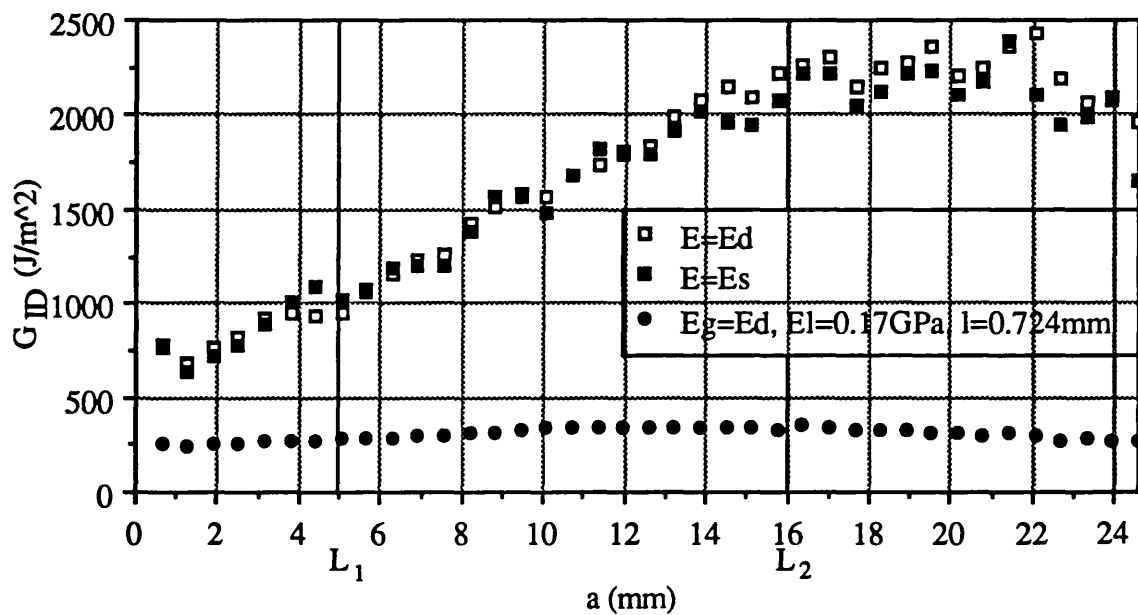


Figure 6.3.4.5: Comparison between G_{ID} results from different FE analyses, test PCSG-2

A good fit between computed and measured peak strains was possible only when a local strip with a small thickness and a very low local modulus - E_l was used. For example, the local strip with $E_l = 0.1$ GPa and $l = 0.567$ mm for the PMMA SG1-1/4 sample, and $E_l = 0.17$ GPa with $l = 0.724$ mm for the glass-PC PCSG-2 specimen, enabled a satisfactory agreement between the experimental and numerical peak strain results (table 6.3.2.1, figures 6.3.4.2-5). When the local strip with a low modulus was employed in the analysis, the predicted variation of the strain (and the stress) near the crack tip with crack length, was much lower than in the single constant modulus analysis. Once the measured peak strains were numerically reproduced, the variation of the calculated crack resistance with crack length was significantly reduced (figures 6.3.4.2-5, table 6.3.4.1). G_{ID} vs a values generated employing a local strip low modulus model are in accordance with the observations detailed in section 6.2. It stated that the crack driving force should be rather insensitive to the crack length for the crack to propagate at constant speed in the SENT sample, leaving a uniform fracture surface behind. This implied that G_{ID} data should be rather independent of the crack length, as obtained by the low local modulus analysis. It is obvious that in the FE simulation of RCP, the accuracy of the generated crack resistance results is critically dependent on the accuracy of the predicted peak strain histories (section 6.3.3, table 6.3.4.1).

Sample	Test case	$\epsilon_{yy2} / \epsilon_{yy1}$	G_{ID2} / G_{ID1}
SG1-1/4	1	1.49	2.
"	2	1.54	2.2
"	3	1.17	1.3
PCSG-2	1	1.3	2.3
"	2	1.2	1.8
"	3	1.04	1.2

Table 6.3.4.1: The peak strains vs G_{ID} predictions, (subscripts 1 and 2 refer to crack lengths L_1 and L_2 respectively; test cases notation 1, 2 and 3 for SG1-1/4 and PCSG-2 corresponds to the notation given in figures 6.3.4.2 and 6.3.4.4 respectively)

When the local strip with a low modulus around the crack tip was employed in the analysis, not only the variation of the calculated crack resistance with the crack length, but also its magnitude, were significantly reduced, compared to the single modulus analysis (figures 6.3.4.3, 6.3.4.5). The comparisons between the changes in energy components, computed by single constant modulus and local strip low modulus analyses, are presented in figures 6.3.4.6 and 6.3.4.7 for SG1-1/4 and PCSG-2 specimen respectively. The local strip low modulus model predicted much larger amounts of retained strain energy and kinetic energy in the samples, than the single modulus approach. Consequently, the low local modulus model generated much less dissipated energy at the crack tip during RCP (figures 6.3.4.3, 6.3.4.5-7). A low modulus in the region close to the moving crack tip resulted in increased inertia effects (Freund and Lee [90], Achenbach et al [79], Broberg[73]). The greater the inertia forces required to move the mass, the larger the retained strain energy in the sample, and the less the energy available to the crack as it passes. This retained energy later became kinetic energy. Although the crack velocity was relatively small compared to the wave speed in the region outside the local strip, it was very close to the longitudinal wave speed in the local low modulus region. Thus, dynamic effects were greatly increased in the local strip low modulus analysis. A more drastic reduction in calculated G_{ID} occurred in the glass-PC than in the PMMA sample (figures 6.3.4.3, 6.3.4.5). During the time interval between the breaking of the glass slide and the crack initiation in the PC, there was an increase in kinetic energy and corresponding decrease in the strain energy, as the flanks of the prenotched crack opened (figure 6.3.4.7). When the local strip low modulus model was used, this amount of kinetic energy, generated in the sample prior to the crack initiation, was significantly increased. As a consequence, there was a large reduction of initial G_{ID} values, compared to the single constant modulus results (figure 6.3.4.5).

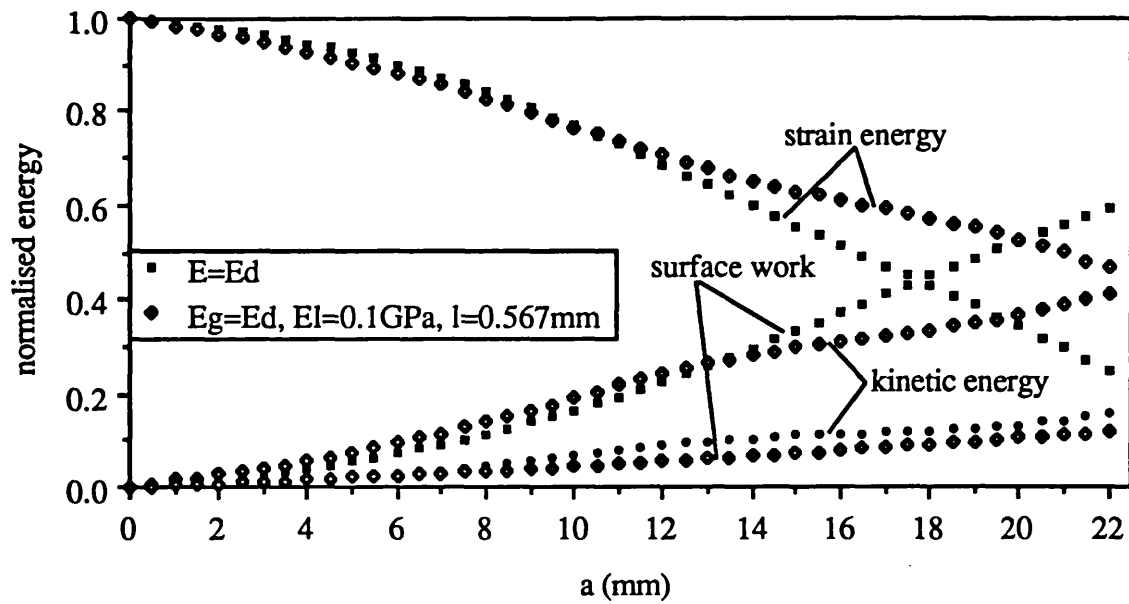


Figure 6.3.4.6: Energy components in SG1-1/4 sample; (normalized by initial strain energy)

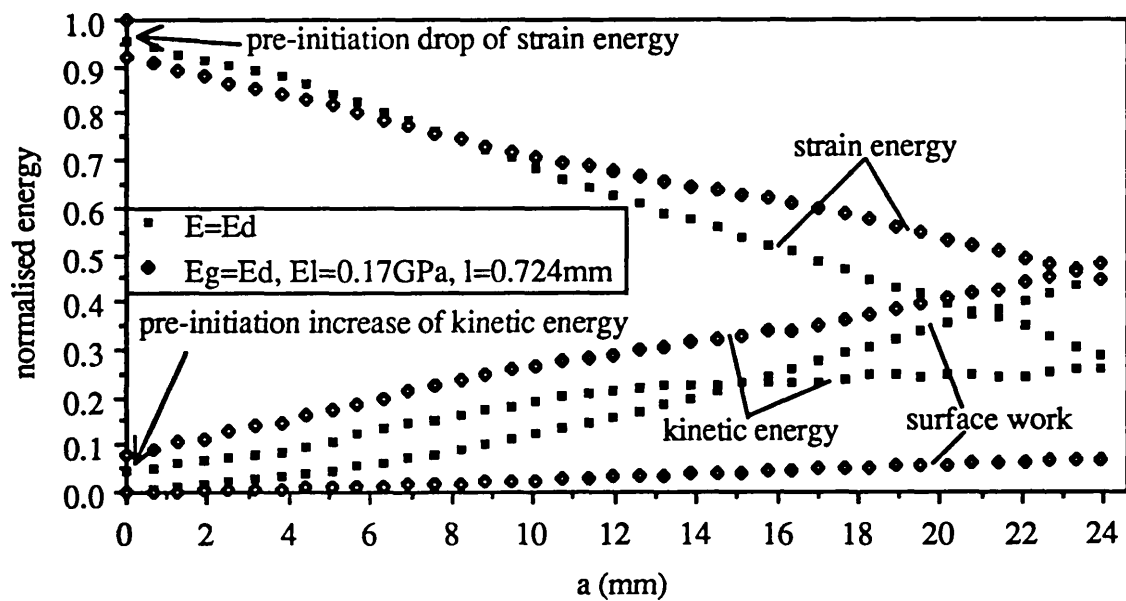


Figure 6.3.4.7: Energy components in PCSG-2 sample; (normalized by initial strain energy)

6.3.5 PMMA SENT summary

Three out of five tested PMMA samples produced records of strain histories from both strain gauges. These samples fractured at different crack speeds: 267, 348 and 364 m/s. The strain records obtained from the SG1 of these three PMMA samples, are presented in figure 6.3.5.1. L_1 and b_1 were approximately the same for all three cases (figure 6.3.2.1, table 6.3.2.1). Figure 6.3.5.1 demonstrates that at a given position near the crack path, the higher peak strain is associated with the lower crack speed in the PMMA samples.

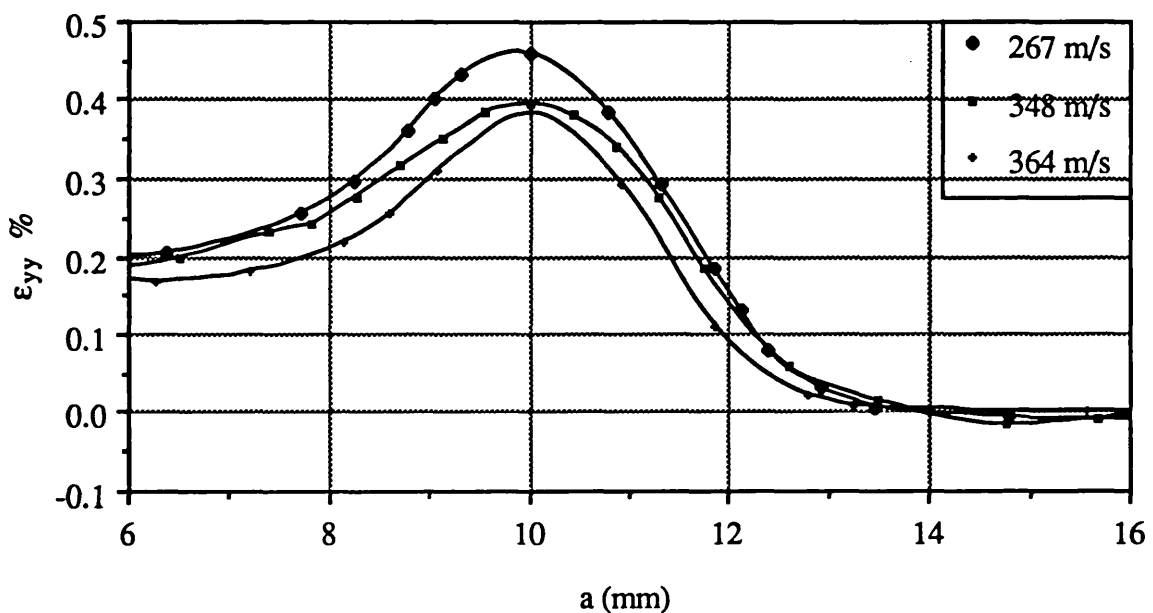


Figure 6.3.5.1: Strain from SG1 vs total crack length, results from three specimens

As demonstrated in the previous section, the local strip parameters (E_1 and l) were determined by fitting the numerical prediction of the peak strains to those measured at the two crack lengths, L_1 and L_2 , during RCP. Such local moduli, obtained from the analysis of the three PMMA samples, are presented in figure 6.3.5.2 as a function of the crack speed. In each

analysed case, an adequate fit was achieved with l between 0.565 and 0.567 mm. The line plotted in figure 6.3.5.2, shows the values of the modulus calculated using equation 1.3.4 and taking the longitudinal wave speed equal to the crack speed. Figure 6.3.5.2 demonstrates that a good agreement between numerical and experimental peak strains was obtained only when the longitudinal wave speed in the local strip region was reduced close to the crack speed. Thus, for a given thickness of the local strip, the higher local modulus corresponds to the higher crack speed (figure 6.3.5.2).

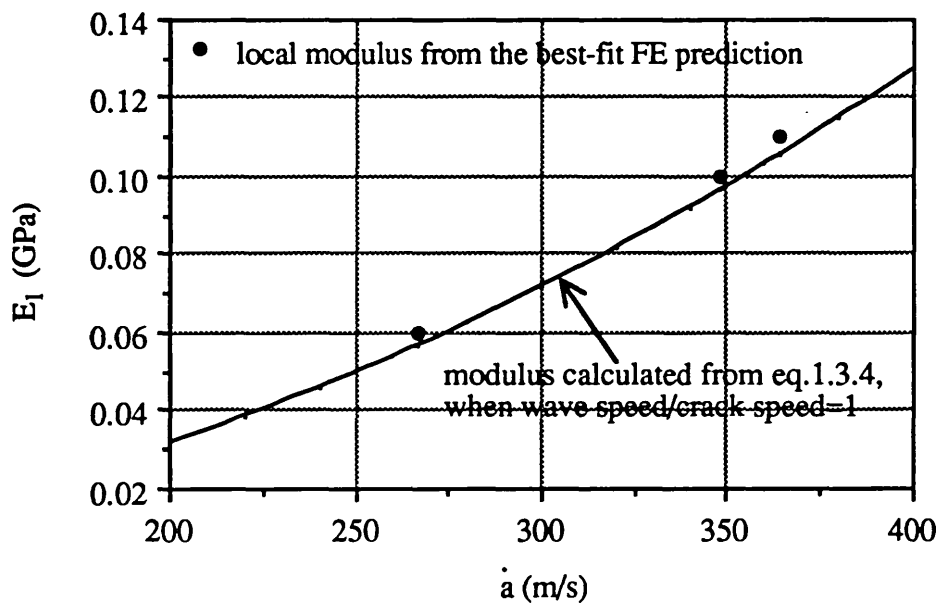


Figure 6.3.5.2: Local modulus vs crack speed

It should be mentioned here that in order to get a satisfactory fit between numerical and experimental results and, thus, to choose the appropriate set of the local zone parameters, a number of sets of various E_l and l were employed. This enables an estimation of the sensitivity of the computed G_{II} results to the variation of local strip parameters (see appendix A3).

The experiments on the PMMA SENT specimens (section 4.4.1) were analysed in section 5.3.3, using a single constant modulus approach. These are re-analysed in this section, using a local strip low modulus model. According to the results presented above (figure 6.3.5.2), the choice of E_1 for a given test was based on the recorded crack speed, i.e. the longitudinal wave speed inside the strip was chosen 3% higher than the measured crack speed, while $l = 0.565$ mm was used throughout. Summarized G_{ID} vs $\dot{a}$ results, from both the single constant modulus (section 5.3.3) and the local strip low modulus analyses, are shown in figure 6.3.5.3. In the latter case, $E_g = E_d$ is chosen for the base material outside the local strip (figure 6.3.4.1, table 2.2.1). Each value of G_{ID} presented in figure 6.3.5.3, was obtained by averaging G_{ID} values from a single test, over the crack length. Due to the uncertainties regarding the crack behaviour at the initiation stage, and the possible inaccuracies developed as the crack approached the end of the specimen, G_{ID} values computed over the first and last five nodal spaces of growth were neglected in the averaging process. This was a rather arbitrary decision.

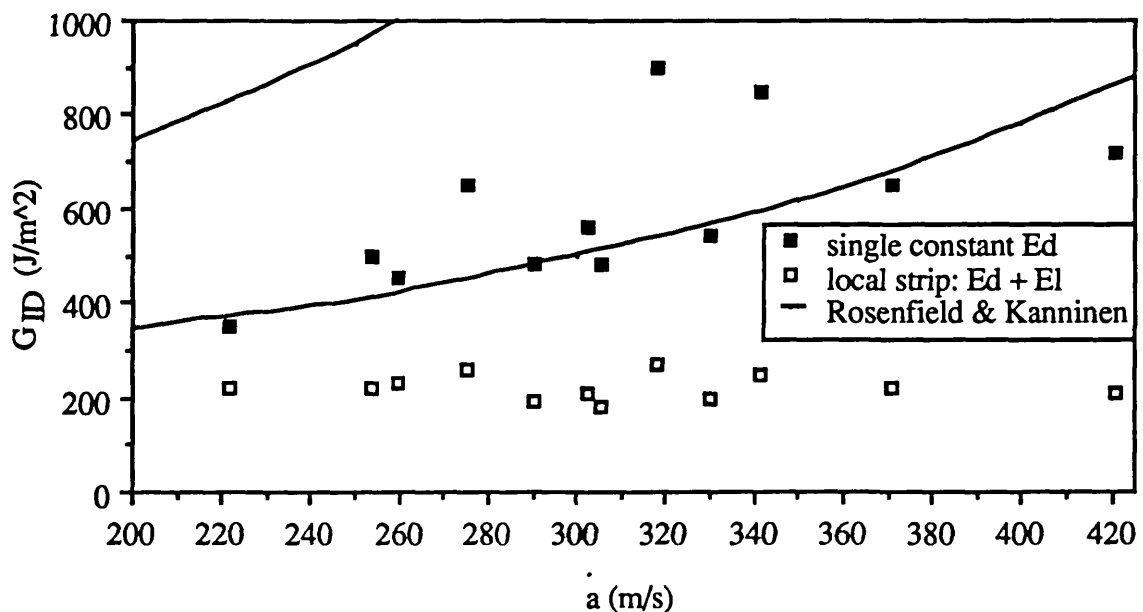


Figure 6.3.5.3: Crack resistance vs crack speed, results of PMMA SENT samples

Also shown in figure 6.3.5.3, are the lower and upper bounds of G_{ID} vs $\dot{a}$ results of PMMA, reported by Rosenfield and Kanninen [73]. They reviewed publications on PMMA and gave a broad band representation of the crack resistance vs crack speed results, within which the results of the majority of workers were claimed to lie. The present work results, generated by the single constant modulus analysis, are also within these bounds (figure 6.3.5.3).

The single constant modulus interpretation of G_{ID} vs $\dot{a}$ results of PMMA, shows highly scattered and multivalued G_{ID} data (figure 6.3.5.3). By including the local strip with a low modulus, these anomalies are essentially eliminated and much of the speed-dependence is removed. The crack resistance data generated by the local low modulus model are nearly independent of the speed and sample size. The average value of G_{ID} is around 220 J/m^2 , and is close to the reported lower bound results at lower crack speeds. This value is in a very good agreement with that recently obtained by Ivankovic and Williams [91]. They analysed RCP data of the PMMA SENT specimens of section 4.4.1 by employing a Mass Spring Model, and reported a crack resistance of about 200 J/m^2 (see appendix A2).

6.3.6 Local low modulus analysis of glass-PC SENT sample

As detailed in section 4.2.2, the glass-duplex SENT sample was used to generate RCP in PC. Genussov[89] reported that the unpolished glass-slide always broke at a very low applied load, failing to generate a fast crack in a test piece. Thus, he suggested the use of a polished glass-slide. On the contrary, when the polished glass-slide was adhered to the PC sample, a very high load was associated with the breaking of the glass-slide. Consequently, RCPs at very high speeds were generated in the PC (section 4.4.2).

Summarized G_{ID} vs crack speed results of PC, from the single constant (E_d) modulus analysis, are presented in appendix A4.

All glass-PC samples involved in the successful measurement of the strain histories in the PC during RCP, fractured at a similar speed of about 530 m/s (section 6.3.2). The local strip low modulus analysis of such glass-PC sample resulted in the longitudinal wave speed in the local strip being close to the crack speed (section 6.3.4). Generated G_{ID} values were nearly independent of the crack length and were approximately 300 J/m² (figure 6.3.4.5). When the PMMA SENT samples were analysed using the local strip low modulus model, the choice of the local parameters was based on the results shown in figure 6.3.5.2 (see previous section). These results are supported experimentally, since they were obtained by fitting the numerical predictions of the peak strains to those measured in the three PMMA samples fractured at different crack speeds (figure 6.3.5.2). Failing to obtain the experimental strain data for a broad range of crack speeds in PC, the determination of E_1 was limited to a single value, corresponding to a crack speed of 535 m/s. Therefore, an assumption of E_1 needs to be made, in order to analyse samples fractured at speeds other than 535 m/s. This assumption was based on the fact that a constant speed RCP is associated with the peak strains and G_{ID} roughly independent of the crack length (figures 6.3.2.3, 6.3.4.4-5). Presented below is the local strip analysis of the glass-PC sample PC4-5, fractured at a speed equal to 406.7 m/s

(section 5.3.4). A value of l equal to 0.724 mm was assumed throughout (section 6.3.4), while E_1 was varied until the above requirements were satisfied (figure 6.3.6.1). $E_g = E_d$ was used outside the layer (figure 6.3.4.1, sections 6.3.3-4).

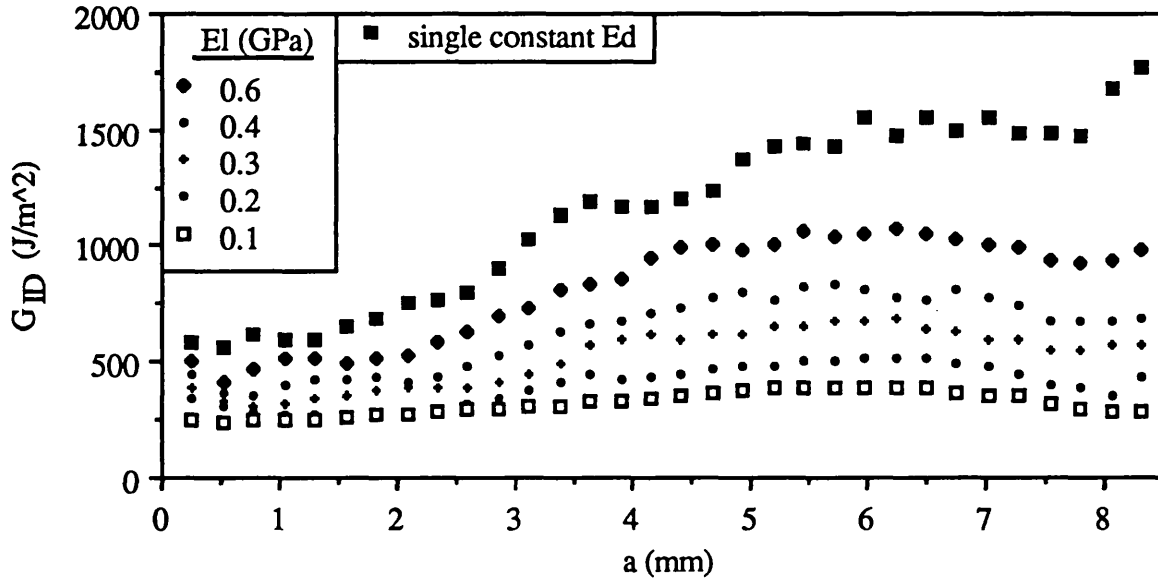


Figure 6.3.6.1: Crack resistance vs crack length increment;
test PC4-5, influence of E_1 on G_{ID} results

Again, when the longitudinal wave speed in the local strip approached the crack speed ($E_1 = 0.1$ GPa), the variation of the local strains and G_{ID} values with the crack length was significantly decreased. The average value of G_{ID} was around 320 J/m² (figure 6.3.6.1). A similar numerical test, performed on a sample PC2-3 which fractured at 648 m/s (section 5.3.4), resulted in $E_1 = 0.26$ GPa and average G_{ID} of 290 J/m². This analysis of the glass-PC SENT samples suggests that the crack resistance of PC is rather independent of the crack speed and is close to 300 J/m².

6.3.7 Local low modulus analysis of duplex PMMA/PC SENT sample

Crack resistance values, generated by the elastic FE analysis of RCPs in the PMMA, glass-PC and duplex PMMA/PC SENT specimens, were sample size dependent and multivalued with respect to the crack length and speed (chapter 5). Since the single constant modulus analysis did not support unique G_{ID} vs crack speed relationships of PMMA and PC, it neither verified the uniqueness of the criterion for the crack propagation across the PMMA/PC interface in the duplex sample (equation 2.4.2.2). When the local low modulus concept was introduced in the analysis, much of the variation of G_{ID} with crack length and speed was removed. Both PMMA and PC were suggested to possess unique G_{ID} vs crack speed values. These crack resistance values of around 220 and 300 J/m² are proposed as material parameters of PMMA and PC respectively. Local strip low modulus analysis of RCP in the duplex PMMA/PC SENT configuration, is therefore expected to generate G_{ID} data around 220 J/m² in the PMMA part, and 300 J/m² in the PC part. Figure 6.3.7.1 presents G_{ID} results obtained from the constant modulus and local strip analyses of the duplex PMMA/PC sample TMN10-2 (table 3.5.2.1). The local strip parameters of both PMMA and PC part of the sample are listed in table 6.3.7.1.

	E_I (GPa)	l (mm)
PMMA	0.09	0.565
PC	0.05	0.724

Table 6.3.7.1: Local Parameters of the PMMA and PC, for the analysis of the duplex PMMA/PC SENT sample TMN10-2

Dynamic moduli are specified for the regions outside the layers. The choice of the local parameters follows from the analyses of the single PMMA and PC samples (sections 6.3.4-6).

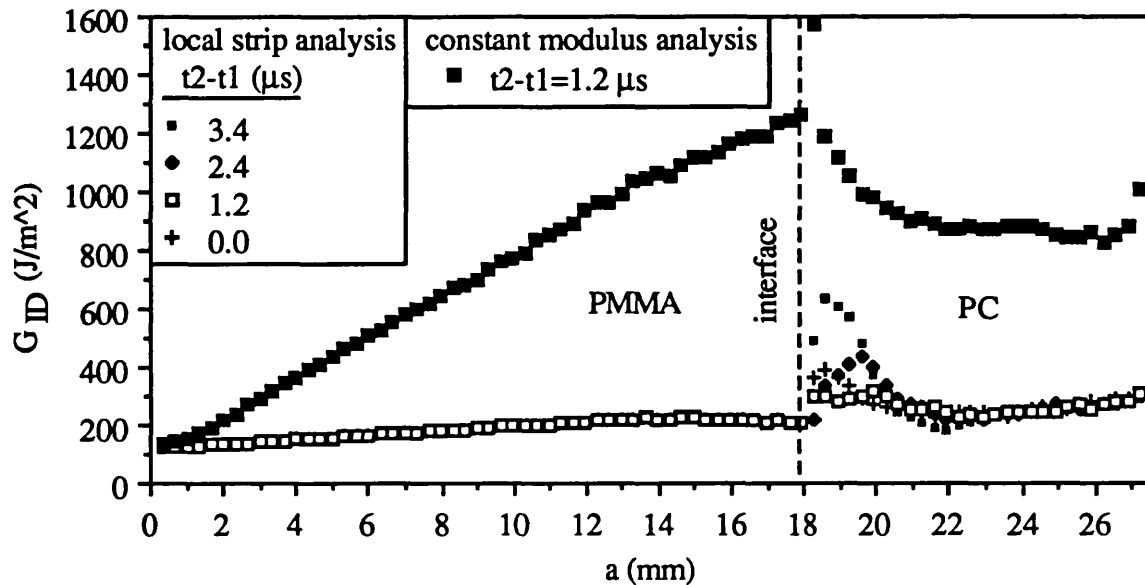


Figure 6.3.7.1: Crack resistance vs crack length increment; constant modulus vs local strip analyses; influence of the duration of the interface crack hesitation on local strip results, test TMN10-2

Figure 6.3.7.1 gives the comparison between G_{ID} results generated by the local strip analysis with various durations (t_2-t_1) of the crack 'hesitation' at the interface (sections 3.5.2, 5.3.2). The experimentally recorded time (t_2-t_1) was 3.4 μs . Away from the interface, G_{ID} data of all local strip tests stabilize around the same value (section 5.3.2). However, the results obtained using $t_2-t_1 = 1.2 \mu s$ correlate best with the fracture surface appearances of the constant speed RCP in the PC (sections 3.5.2, 6.2, 6.3). As shown in section 3.5.2, a crack entered the PC part of the duplex sample with a front slightly curved near the sample sides. Consequently, the recorded time (t_2-t_1) was an overestimation of the true crack 'hesitation' time (section 3.5.2),

as suggested by the local strip analysis. The most important outcome of the local strip analysis of the duplex PMMA/PC SENT sample is that G_{ID} values of both the PMMA and PC parts are remarkably close to those obtained from the single material SENT geometry. This strongly suggests the existence of a unique, geometry independent G_{ID} of the PMMA and PC, and demonstrates the validity of the criterion for the crack propagation across the PMMA/PC interface in the duplex sample (equation 2.4.2.2).

SUMMARY and RECOMMENDATIONS

The work described represents an investigation of RCP in PMMA, PC, PVC and two-layer PMMA/PC and PMMA/PVC systems. This study involves: i) the analysis of the crack behaviour in polymer brittle/ductile systems, ii) the characterization of the dynamic crack resistances of PMMA and PC, iii) the investigation of the influence of specimen geometry and size on the RCP properties of the tested materials, and iv) an attempt to solve the problem of geometry dependent and multivalued crack resistance data with respect to the crack speed, as generated by elastodynamic FE code.

RCPs were monitored using three different specimen configurations. The duplex SENT configuration was used for investigating RCP in PMMA/PC and PMMA/PVC systems, SENT geometry was employed for PMMA and the glass-duplex configuration was used for PC and PVC. Each specimen was clamped rigidly between two heavy grips and loaded slowly in tension until RCP occurred. Heavy clamps ensured fixed-grip conditions while the crack propagated. The testing was performed at room temperature. Crack histories were recorded by the high speed photography and the timing lines techniques. The timing line method required a longer time for the preparation of samples, but it proved to be more reliable than the fast photography method. Major problems encountered during the experimental part of this work are: i) the difficulties in setting up of the recording equipment when the high speed photography method was used, ii) the inability to monitor the crack behaviour at the initiation stage when either technique was used, and iii) the difficulties in controlling the glass breaking event when the glass-PC and glass-PVC samples were tested. The last problem might be overcome by applying a 'light' impact to the glass-slide (e.g. with a hammer), at the moment when a desirable preload is reached (see Genussov[89]).

Remarkably constant crack velocities were observed in both PMMA and PC, regardless of the specimen configuration. These constant speed RCPs were associated with uniform fracture surfaces. Also, higher crack speeds were found to correspond to higher preloads. RCP in the

PMMA/PC sample was characterized by a higher crack speed in the PMMA than in the PC section. Contrary to the results published by Rolland[83, 85], no critical speed was observed to correspond to the crack propagation across the PMMA/PC interface. Neither the glass-PVC nor duplex PMMA/PVC SENT geometry generated the required mode of failure in PVC. Either extremely rough fracture surfaces or crack branching occurred in PVC when the glass-duplex configuration was used. The crack branching always occurred in the PVC part of the duplex PMMA/PVC sample. One possible way of generating RCP with a uniform planar fracture surface in PVC, is to use the glass-duplex configuration and provoke the breaking of the glass-slide at a low preload. The main characteristic of RCP in the duplex samples is that an unstable crack was promoted in PC and PVC parts of slowly loaded samples by either the fast crack in the PMMA part of the duplex PMMA/PC and PMMA/PVC samples, or by sudden breaking of the glass-slide of the glass-PC and glass-PVC specimens.

An elastodynamic, node-release FE code was used to analyse the experimental RCP results. The program was designed to model symmetrical bodies of isotropic linear elastic materials. Constant strain triangular elements were used in this two-dimensional code and the sample was represented by an FE mesh gradually refined towards the crack path. Because of symmetry only half of the sample was modelled and the crack path became a boundary. The crack was propagated by the sequential release of crack path nodes. The calculation of the crack resistance was based on the the recorded crack length histories, the magnitude of the breaking load and a single dynamic modulus. The crack extension criterion was based on the global energy balance which assumes that the equality between the crack driving force and the crack resistance exists as long as the crack propagates. This criterion was offered as the crack extension criterion in the duplex samples as well. The single constant modulus analysis of RCPs at constant speeds generated geometry dependent and multivalued crack resistance data with respect to the crack speed. The very large variations of the computed G_{II} with crack length did not correspond to the fracture surface appearances. This outcome of the elastic analysis of the experimental RCP results was found to be unacceptable.

It was concluded that the crack driving force should be independent of the crack length for the crack to advance at a constant speed leaving behind a uniform fracture surface. The accuracy of the FE prediction of the strain variation with crack length during RCP, was examined by measuring the strain histories from two strain gauges adhered close to the crack path. When single constant modulus analysis was performed, the numerically predicted variation of the peak strains along the crack path during RCP was significantly larger than the measured variation. This was found to be responsible for the large variations and errors in the calculated crack resistance vs crack length (and speed). At least two, near crack path strain histories have to be recorded and some kind of full dynamic analysis employed if the validity of predicted strain histories are to be assessed. The improper representation of the material behaviour near the moving crack tip was suggested as the major error in the single constant modulus analysis of RCP in tested materials. Since the parameters describing such behaviour are unknown, the local strip low modulus model was used to approximate eventual non-linearities of the material around the crack tip. The local low modulus model predicted much larger amounts of retained strain energy and kinetic energy in the samples, than the single constant modulus approach. Consequently, energy dissipated at the crack tip during RCP was significantly reduced. Once the recorded peak strains were reproduced, the variation of G_{ID} with crack length was considerably decreased. G_{ID} data of PMMA and PC generated by the local low modulus model were nearly independent of the crack speed and around 220 and 300 J/m² respectively. When this model was applied to the analysis of RCP in the duplex PMMA/PC sample, G_{ID} values of both the PMMA and PC are reproduced. It was suggested that both PMMA and PC possess unique, geometry independent G_{ID} values. Also, the validity of the energy balance criterion for RCP in the duplex sample was demonstrated.

The representation of the material behaviour near the advancing crack tip was shown to be critical in the analysis. Future workers in this field are therefore strongly recommended to calibrate the material mechanical properties under the conditions which correspond to those near the moving crack tip. For this purpose, strains, strain rates and temperature have to be determined in the vicinity of the crack tip during RCP. After this, a more realistic modelling of RCP, could be carried out.

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Appendix A1

Stress wave propagation in non-linear material

The example presented below is chosen because of its attractive simplicity. It provides an insight into the problem of the stress wave propagation in an 'environment' with non-linear material behaviour. Consider a bar of idealized material having a bilinear stress-strain relationship (figure A1.1), such that:

$$E = E_1 \quad \text{for } \sigma \leq \sigma_1,$$

$$E = E_2 \quad \text{for } \sigma > \sigma_1$$

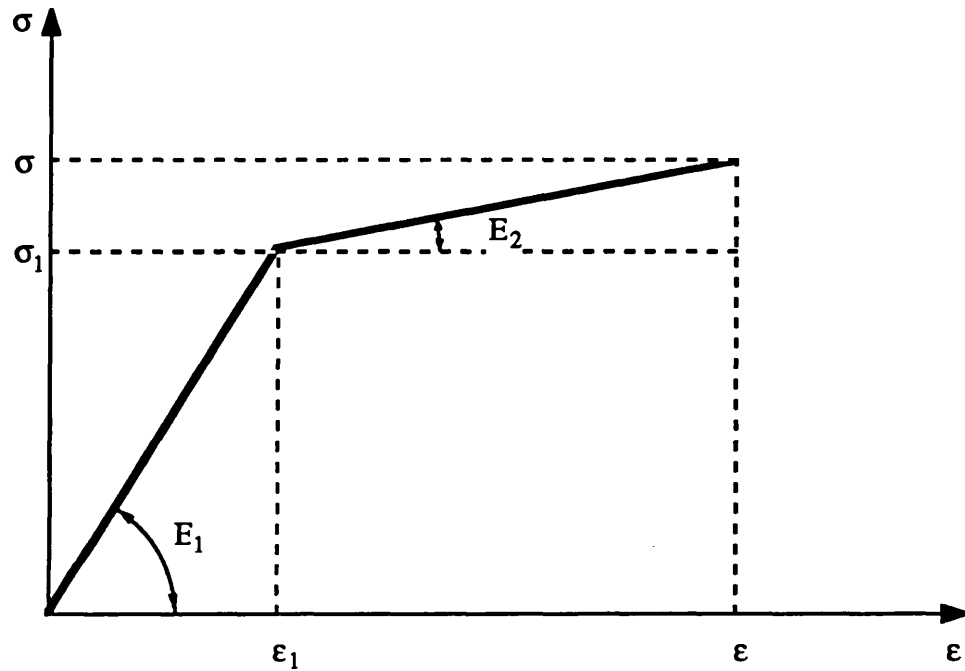


Figure A1.1: Stress-strain curve for bilinear material

When the nominal stress $\sigma \leq \sigma_1$ is suddenly applied at the end of the bar, the stress wave will start to propagate through the unstrained bar at a speed of $(E_1/\rho_0)^{1/2}$. The bar is assumed to be

infinitely long, thus the stress wave reflection is avoided. Further, if a stress $\sigma > \sigma_1$ is applied so that the stress increment $(\sigma - \sigma_1)$ and strain increment $(\epsilon - \epsilon_1)$ are related through E_2 , then it seems that this excess stress should be propagated at a speed of $(E_2/\rho_0)^{1/2}$ through the bar. Assume that the bar is loaded to a nominal stress level of $\sigma > \sigma_1$ instantaneously. A strain ϵ is associated with the applied stress (figure A1.1). From the loaded end of the bar, this stress is transmitted through the bar by two waves which start at the same instant and move at the different speeds: $C_1 = (E_1/\rho_0)^{1/2}$ and $C_2 = (E_2/\rho_0)^{1/2}$. The stress and strain distribution at time t after loading, is shown in figure A1.2. As t increases the distance between the fronts of the waves is increasing as well.

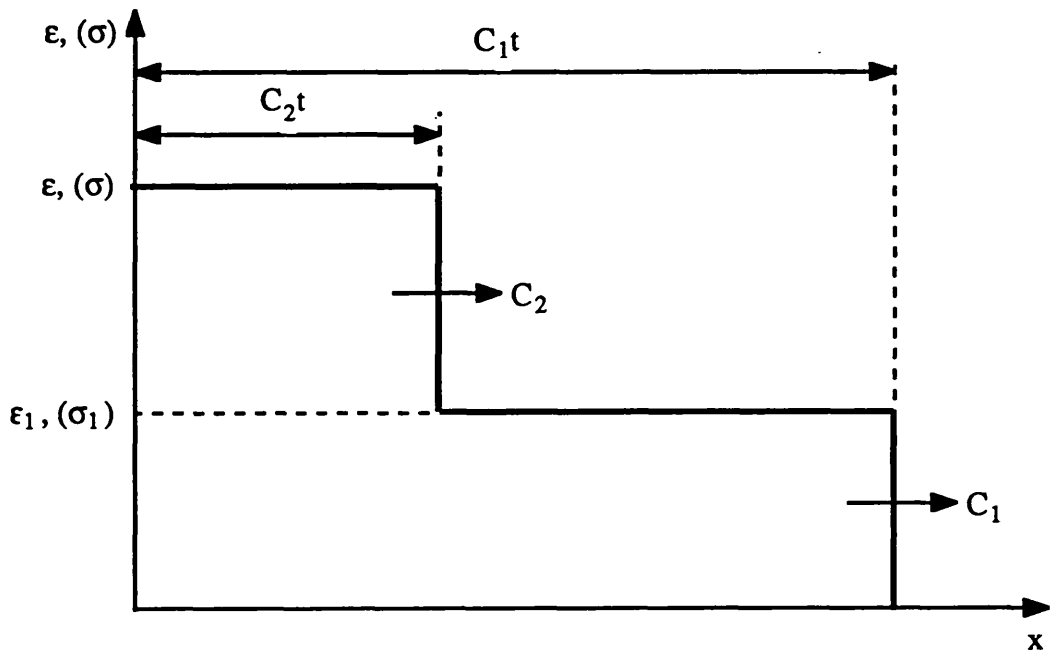


Figure A1.2: Waves propagation through the bar with bilinear stress-strain relationship

Three distinct regions at a given time t can be identified (by reference to the impacted end of the bar at time $t = 0$, figure A1.2):

- $0 < x < C_2t$, the strain is equal to ϵ and the stress wave is propagating at C_2 ,
- $C_2t < x < C_1t$, the strain is equal to ϵ_1 and the stress wave is propagating at C_1 ,
- $x > C_1t$, the bar is unstressed ahead of the wave front.

Appendix A2

Mass-Spring Modelling

The simple tension test may be modelled by a two spring-single mass model shown in figure A2.1. k_1 is the contact stiffness comprises the grips and machine stiffness, and $k_2(a)$ is the crack length dependent specimen stiffness. m is the equivalent mass of the specimen.

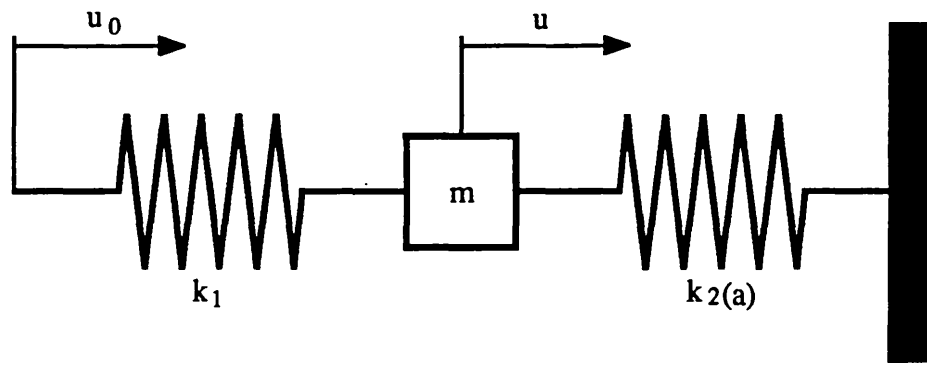


Figure A2.1: The mass-spring model

The mass of the clamps is not included here since measurements of the machine and clamp stiffness showed that they are essentially immobile and all the kinetic energy arises in the specimen. The applied displacement (u_0) gives rise to a displacement in the specimen (u) of:

$$u = \frac{k_1}{k_1 + k_2} u_0 \quad \text{A2.1}$$

In the analysis, it is assumed that all motion arises from the crack growth and that u_0 is constant. The energy release rate (G), for the short crack lengths, is given by Williams[87] as:

$$\frac{G}{G_s} = 1 - \frac{1}{C^2} \frac{d(a \dot{a})}{dt} \quad \text{A2.2}$$

where:

$$G_s = \left[\frac{\left(\frac{k_1}{BE}\right) H}{D + \left(\frac{k_1}{BE}\right) H} \right]^2 \left(\frac{u_0}{H}\right)^2 E \pi a \quad A2.3$$

is the static energy release rate and $\bar{C}$ is a characteristic speed given by:

$$\bar{C} = C \sqrt{\frac{3}{4\pi}} \left[\sqrt{\frac{D}{H}} + \left(\frac{k_1}{BE}\right) \sqrt{\frac{H}{D}} \right] \quad A2.4$$

and:

$$C = \sqrt{\frac{E}{\rho}} \quad A2.5$$

the tensile wave speed.

$\bar{C}$ depends on the material properties mostly via C but is also geometry dependent via D , H and B , and is system dependent via k_1 . It is significant that a minimum occurs in $\bar{C}$ when $D/H = k_1/BE$ and that $\bar{C}$ varies rather slowly with D/H .

If we define a G_s value for $a = a_{0t}$ as G_0 ($a_{0t} = a_0 + a_1$, section 4.2.1), and assume $G = R$, a constant crack resistance, then equation A2.2 becomes an equation of motion for a :

$$\frac{R}{G_0} \frac{a_{0t}}{a} = 1 - \frac{1}{\bar{C}^2} \frac{d(a \dot{a})}{dt} \quad A2.6$$

which may be solved for the boundary conditions $\dot{a} = 0$ at $a = a_{0t}$ to give:

$$\left(\frac{\dot{a}}{\bar{C}}\right)^2 = \left(1 - \frac{a_{0t}}{a}\right)^2 + 2 \frac{a_{0t}}{a} \left(1 - \frac{a_{0t}}{a}\right) \left(1 - \frac{R}{G_0}\right) \quad A2.7$$

In the real experiment an apparently almost constant speed over much of the crack growth was observed. For simplicity, we shall assume that the speed was observed at, say, half way across the ligament, i.e. at:

$$a = \frac{D + a_{0t}}{2} \quad \text{A2.8}$$

from which:

$$\frac{a_{0t}}{a} = \frac{2 \frac{a_{0t}}{D}}{1 + \frac{a_{0t}}{D}} \quad \text{A2.9}$$

In all tests (section 4.2.1), $a_{0t} = 0.2D$ so $a_{0t}/a = 1/3$, and from the equation A2.7:

$$\frac{1}{EG_0} = \frac{2}{ER} - \frac{9}{4ER} \frac{\dot{a}^2}{\bar{C}^2} \quad \text{A2.10}$$

Further, G_0 can be expressed as:

$$G_0 = \frac{\pi \sigma^2 a_{0t}}{E}$$

where, the nominal stress σ is:

$$\sigma = \frac{P_b}{BD}$$

and P_b is the experimentally recorded breaking load. Thus, the left hand side term of equation A2.10 takes the form:

$$\frac{1}{EG_0} = \frac{1}{\pi \left(\frac{P_b}{BD}\right)^2 a_{0t}}$$

and is calculated for each test from section 4.4.1.

Figure A2.2 shows $1/EG_0$ as a function of the square of measured crack speed ($\dot{a}^2$) for various D/H values. The lines (figure A2.2) are fitted through each set of points, i.e. for each D/H ratio. From the intercepts between these lines and the horizontal axes, $\bar{C}$ data are obtained for

different D/H values using equation A2.10. A crack resistance (R) of 200 J/m^2 is generated from the common intercept of the lines and vertical axes. This value is very close to that of 220 J/m^2 generated by the local low modulus model (section 6.3.5).

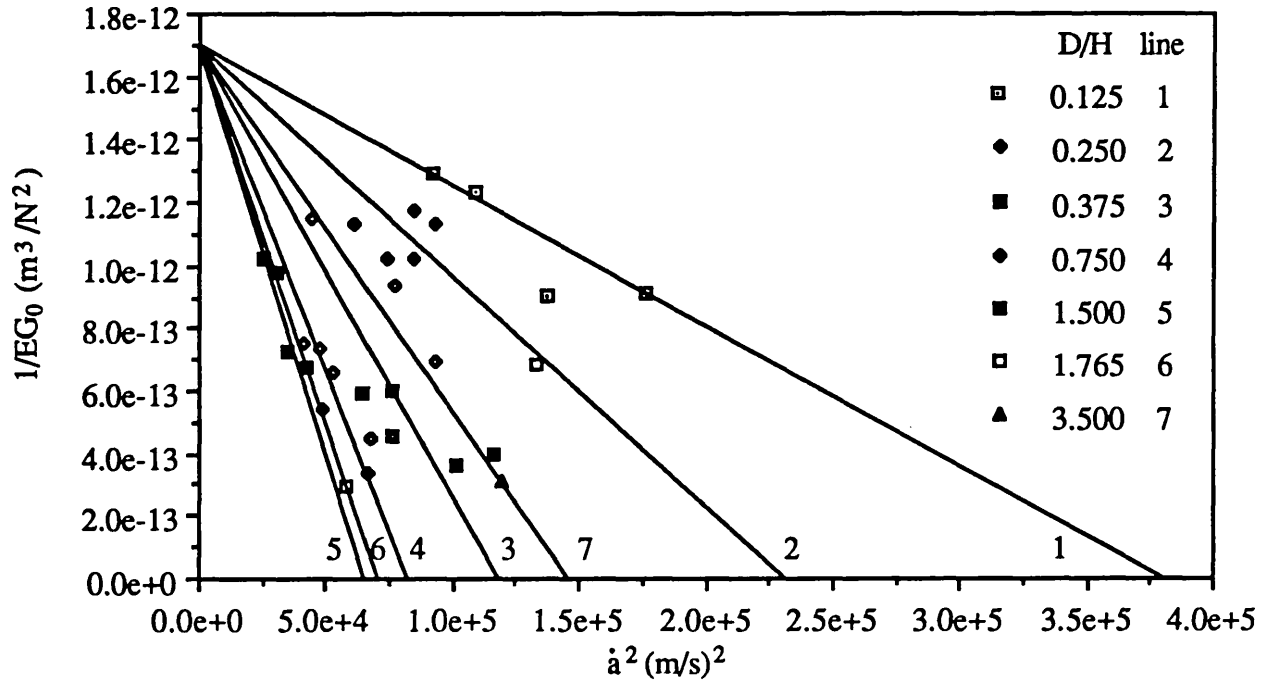


Figure A2.2: $1/EG_0$ vs $\dot{a}^2$, experimental results

Figure A2.3 presents the $\bar{C}$ data as a function of D/H ratio. The line, drawn by employing the equation A2.4 and using measured $k_1 = 1 \cdot 10^8 \text{ N/m}$ with the global modulus $E = E_d$, is also shown in figure A2.3. The minimum in $\bar{C}$ occurs for $D/H = 1.5$. The most important observation is the very good agreement between the experimental $\bar{C}$ data and theoretical prediction (equation A2.4) with $C = 230 \text{ m/s}$. This is much lower value of C than would be expected (2745 m/s) and it gives the modulus of $7 \cdot 10^7 \text{ N/m}^2$, which is close to E_1 used in the FE analysis.

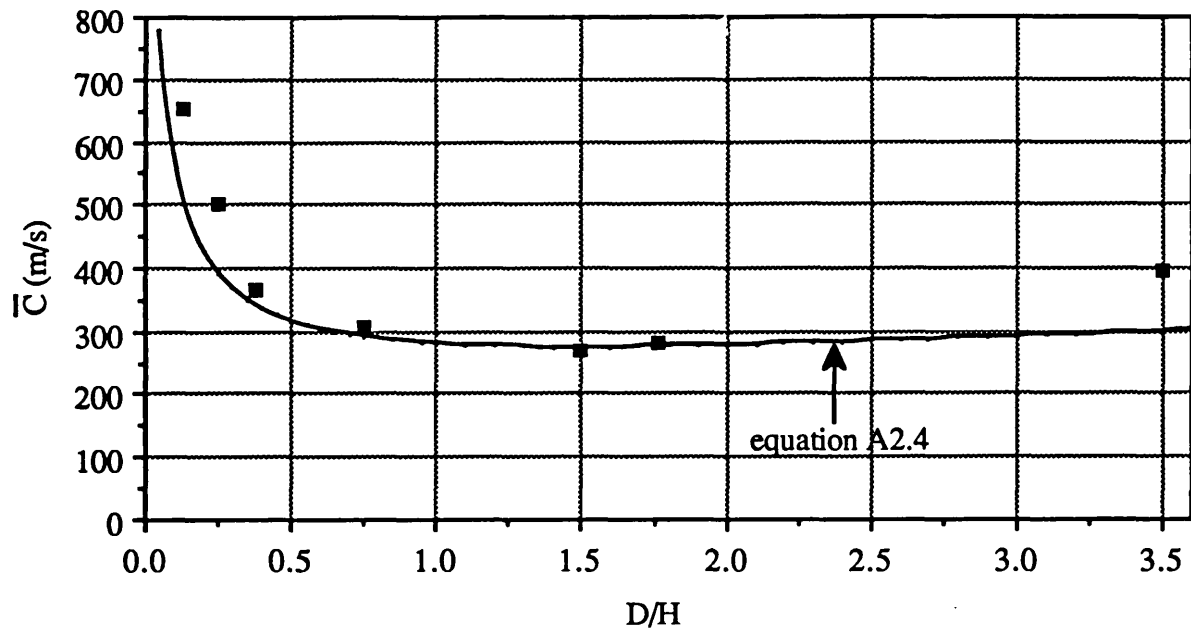


Figure A2.3: $\bar{C}$ vs D/H

Appendix A3

Influence of the local strip parameters on G_{ID} results

A number of sets of various local strip parameters (E_l , l) was employed in the analysis, in order to adequately reproduce experimental peak strains (section 6.3.4-7). The sensitivity of the computed G_{ID} results to the variation of E_l and l is illustrated in figures A3.1 and A3.2 respectively. Presented results are from the analysis of the test SG1-1/4 (section 6.3.4). The minimum value of l was restricted by the size of the elements around the crack path in the FE mesh (l cannot be smaller than the element height). Numerical instability occurred when the crack speed exceeded the longitudinal stress wave speed in the local layer about twice. This limited the minimum value of E_l . At least two rows of elements were used to represent one half of the local strip (due to the symmetry only half of the sample was modelled). For a given l , the average value of G_{ID} and its variation with the crack length decrease as E_l decreases (figure A3.1). On the other hand, for a given E_l , the average value of G_{ID} as well as its variation with the crack length increases with decreasing l (figure A3.2). Eventually, G_{ID} should approach the single constant modulus results as l approaches zero (Broberg[73]). Note that nearly the same G_{ID} results are obtained by employing different sets of the local strip parameters, i.e. $E_l = 0.05$ GPa with $l = 0.433$ mm produces the same G_{ID} values as $E_l = 0.1$ GPa with $l = 0.567$ mm (figure A3.2). However, the local strip model with $E_l = 0.05$ GPa and $l = 0.433$ mm, did not match the experimental peak strains. The best fit between numerical and experimental peak strains was obtained with $E_l = 0.1$ GPa and $l = 0.567$ mm (section 6.3.4).

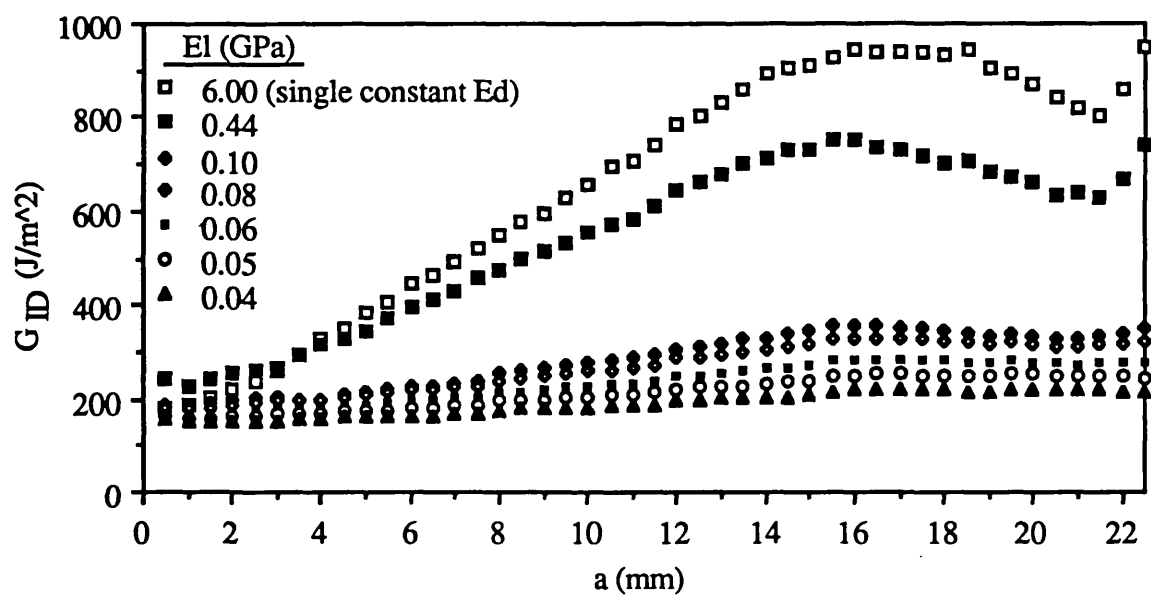


Figure A3.1: Influence of E_1 on G_{ID} results; $l = 0.433$ mm, test SG1-1/4

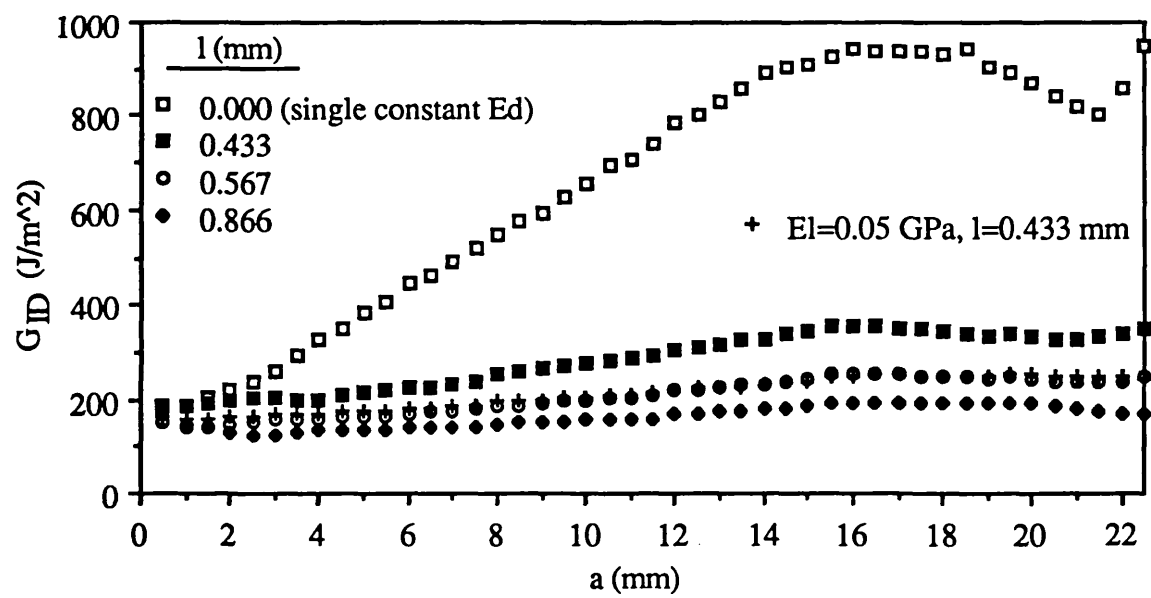


Figure A3.2: Influence of l on G_{ID} results; $E_1 = 0.1$ GPa, test SG1-1/4

Appendix A4

Summary of G_{ID} results of PC

G_{ID} vs crack speed results of PC, from the single constant dynamic modulus analysis of both duplex PMMA/PC and glass-PC SENT configurations (sections 5.3.2 and 5.3.4), are shown in figure A4.1. Each G_{ID} value presented in figure A4.1, was obtained by averaging G_{ID} values from a given test over the crack length (see section 6.3.5).

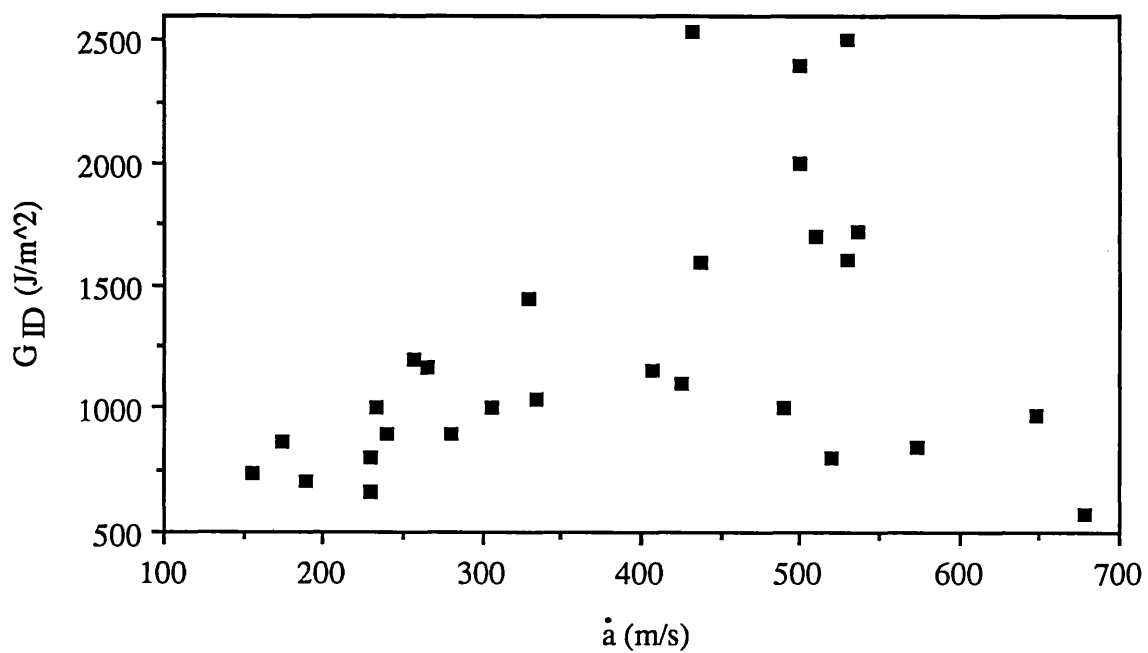


Figure A4.1: Crack resistance vs crack speed, PC results